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THE METALLURGICAL DESIGN OF STEELS FOR OPTIMUM MECHANICAL PROPERTIES

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22 OCT. 1945 (11) 22 Oct 45

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Watertown Arsenal Laboratory
Report Number WAL 320/28
Problem Numbers H-5.1 and H-9.1

22 October 1945

THE METALLURGICAL DESIGN OF STEELS FOR OPTIMUM MECHANICAL PROPERTIES

The primary metallurgical problem connected with guns, armor, projectiles, and many other ordnance parts is the design of steels and heat treatments that will produce the optimum combination of mechanical properties. The severe service that these ordnance parts encounter requires that they have the maximum toughness consistent with the desired hardness. It is now possible to establish a design procedure that will insure the optimum combination of mechanical properties for these parts. However, many of the data and principles for performing the actual design have not been established. In this report the known information is reviewed and assembled, and the deficiencies in the knowledge are pointed out. From this assembled information scientifically sound metallurgical design principles are established. An example utilizing these principles is also presented.

The inspiration for this report came from Col. H. H. Zornig, the former director of the Watertown Arsenal Laboratory. It was his vision that metallurgy could be made an exact science and its application to the design of products and processes could be made quantitative. Many people have aided the authors in preparing this report. They wish to acknowledge particularly the assistance and constructuve criticisms of Dr. E. C. Bain, Dr. M. Grossmann, Capt. S. Herres, Mr. J. Hodge, Dr. H. H. Lester, Dr. J. Low, Miss M. Norton, Lt. Col. N. A. Matthews, Mr. R. F. Miller, and Dr. C. Zener. The photomicrographs used in this report were prepared by Miss M. Norton. For assistance in preparing the manuscript they wish to thank B. Burghardt, V. Civetti, W. Clancy, T. Curley, H. Forristall, S. Palashis, and H. Sturtevant.

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INTRODUCTION

During the many thousands of years that metals have been used to supply the needs and comforts of man, he has based his choice of metal for each use upon past experience with the article to be made. Prehistoric man made crude weapons as his father did. Man even now chooses metals by trial and error and by experience. Many applications require a choice between different metals, while for others a choice between varieties of a single type of metal must be made.

Steel is frequently chosen for many metal parts on the basis of cost, strength, strength-weight ratio, and versatility. All applications of steel, however, do not require the same combinations of properties. Frequently strength alone is needed; in other cases, strength combined with machinability is required. For many steel parts the optimum combination of strength, life, and toughness is desired while high-temperature and corrosion-resistant properties are not needed. Such parts form the largest field of application for so-called "medium alloy steels".

Research carried on in the last few years has so far advanced the science of ferrous physical metallurgy that the selection of steels for this type of part can be made on a basis broader than that of experience and by methods other than trial-and-error. Sound technical principles have been, and are being, developed that permit a rational metallurgical design of steels for parts to have the optimum combination of mechanical properties. One of the principles which has been established and upon which a design method can be based is that the ordinary mechanical properties of steels - its tensile and yield strengths, elongation, reduction of area, fatigue properties, and energy absorption in notched-bar impact tests - are determined primarily by the microstructure of the steel. (This is true also of those more fundamental variables, the flow and fracture stresses under various conditions of temperature, rate of loading, combined stress, and deformation.) Composition affects these properties significantly only insofar as it affects the microstructure. The microstructure is determined by the carbon content and by the phase transformations which the steel undergoes during heat-treatment (either intentional or incident to casting and working), and to some extent, during deformation. The various alloying elements have little effect upon the nature of the transformations, but they do influence the temperatures at which the transformations take place, and markedly affect the speed of the transformations.

Thus, in order to obtain desired mechanical properties, it is necessary only to produce in the steel a definite microstructure. To produce this microstructure, a carbon and alloy content must be chosen that will permit the necessary transformations with the selected heat treatment. The heat treatment must consist of heating, holding, and cooling cycles that will permit these transformations to take place with the selected composition. In the selection of both heat treatment and composition, freedom from various metallurgical and process difficulties, as well as cost and availability of materials and equipment, must be considered.

It has become rather widely recognized that the optimum combinations of strength, life, and toughness can be obtained with steels having a microstructure of tempered martensite. The following chapters deal with the fundamentals and procedures of the metallurgical design necessary for obtaining this microstructure throughout steel parts of uniform composition. There are, in principle, three branches of such design: design of the shape and mechanical properties, design of the composition, and design of the heat-treatment. That the design problem consists of three interrelated parts is not always recognized. All too often, the shape and mechanical properties are designed solely on the basis of the functions of the finished part (with no consideration of metallurgy), the composition is picked from a handbook or stock list, while the choice of heat-treatment is left wholly to the foreman of the heat-treatment shop involved in production of the piece. Such procedures have undoubtedly been successful in many cases, but in others they have caused unnecessary grief and expense. Even metallurgists have been inclined to consider design of shape and properties, design of composition, and design of heat-treatment as three separate problems to be handled in that order, rather than a single problem with three interrelated parts. If there is to be an order, it would seem that it must be: design of shape and properties, design of heat-treatment, design of composition. The heat-treatment (at least the quenching) is determined primarily by what can be done to the piece without cracking or excessive distortion, while the composition is determined by the requirement that the specified properties must be produced in the specified shape with the selected heat-treatment. However, many details of the heat-treatment are affected by the composition, while both heat-treatment and composition affect the extent to which unequal section thickness and re-entrant angles can be permitted in the shape. Shape and properties, heat-treatment, and composition must be designed, in principle, simultaneously. This may not be altogether practical, but it is necessary when undertaking one of these branches of metallurgical design at least to keep in mind the other two. Even if, as seems best, shape and properties, heat-treatment, and composition are designed in that order, it will frequently be necessary, when considering the composition, to modify the heat-treatment, while after considering heat-treatment and composition, it may be necessary to modify the shape and properties. No one of these branches of the design can be completed until the other two are also completed.

Before metallurgical design for tempered martensite can be undertaken, it is necessary to understand how tempered martensite is formed. To obtain in a steel this structure, it is necessary first to obtain the structure martensite, and then to temper it. Martensite is obtained by heating steel above a critical temperature range and holding for a suitable time, so that the structure austenite forms, and then quenching to a temperature low enough to transform the austenite to martensite, and rapidly enough to prevent the formation of other structures. The purpose of the austenitizing and quenching operations, which together may be referred to as "hardening", is simply to produce martensite. Martensite is tempered by heating below the critical temperature range, but ordinarily above room temperature, and holding for a suitable time, so that the martensite transforms to what is best termed simply "tempered martensite",

then cooling to room temperature. The purpose of the tempering operation is to produce the desired mechanical properties, together with freedom from internal stresses. The hardening and tempering operations thus consist of a series of temperature-time cycles which successively produce austenite, transform the austenite to martensite and not to any other structure, and transform the martensite to the desired tempered structure.

Three types of basic knowledge are, therefore, involved in metallurgical design for tempered martensite. One concerns the transformations that occur in steels of various compositions when subjected to various temperature-time cycles, and is a problem in kinetics of reactions. Another deals with the methods of producing desired temperature-time cycles in steel parts of various shapes, and is a problem in heat flow. And thirdly, an understanding of the mechanical behavior of steel is required for intelligent interpretation of the behavior of the finished part, as well as intelligent testing of the steel in the various stages of processing.

The three subjects, transformations, heat flow, and mechanical behavior, include the fields of knowledge upon which metallurgical design of heat-treated parts must be based and which are prerequisite to the engineering information for design. The first part of this volume discusses these fundamentals. Next, engineering principles and data are presented which the metallurgist can use in designing the combination of heat-treatment and composition that will produce optimum properties in a part. Separate chapters are devoted to the subjects of mechanical properties, quenching, hardenability, quench-cracking, and temperability.

In the first of these chapters, the significance of the strength, ductility, hardness, impact, fatigue, and creep properties is discussed on the basis of the more fundamental characteristics of the mechanical behavior, the flow and fracture strengths. The relations of these properties to the microstructure are brought out, as well as in certain cases, the effect upon them of temperature, rate of deformation, and combined stresses. The tests used to measure these properties are also described.

The chapter on quenching covers the characteristics of the various practical quenching media and the size relations among the various shapes having equivalent minimum cooling conditions. These relations are derived from the principles discussed in the earlier chapter on heat flow, and are very useful in metallurgical design. The chapter on hardenability discusses the effects of alloying elements upon the pearlite and bainite transformations, which must be avoided during continuous cooling in order to form martensite. Methods of estimating the hardenability of steel from chemical composition are discussed and the deficiencies in the presently available methods are pointed out.

As alloying elements are added to steel to increase its hardenability, or carbon is added to increase its strength, the tendency toward quench-cracking increases. Frequently, therefore, the design of the heat-treatment-composition combination must be adjusted to minimize this difficulty. In some cases, the part is so prone to cracking that changes must

be made in its shape. A separate chapter is devoted to a discussion of the methods of minimizing quench-cracking. Adjustments of the heat-treatment are considered, and the consequently necessary changes in composition are discussed. General principles for choosing compositions to minimize quench-cracking while maintaining adequate hardenability are also presented.

A quenched steel part must almost always be tempered to reduce the quenching stresses. Tempering also permits closer control of the final strength of the steel, and allows subsequent hot fabrication, such as welding or straightening, without softening the part. Tempering must be done at a temperature sufficiently high to accomplish the purpose but not so high as to destroy the martensitic structure. The tempering cycle must be such as to minimize temper brittleness. In choosing a composition, therefore, the response of the steels to tempering must be considered, and a chapter is devoted to this subject.

Based fundamentally upon the principles of transformation, heat flow, and mechanical behavior outlined in the early chapters, and immediately upon the engineering data and principles next presented, a method is finally developed for the metallurgical design of parts to have a tempered-martensitic structure. Briefly, this method involves first the tentative choice of a quenching procedure suitable for the shape to be heat-treated. Next, the hardenability required to harden the piece with the tentative quenching procedure is estimated, and, if desired, is checked. The carbon content is selected tentatively as well as percentages of the various alloying elements that give promise of meeting the estimated requirements for both pearlitic and bainitic hardenability, together with, if desired, minimum tendency toward quench-cracking and retained austenite difficulties. After a suitable austenitizing procedure has been found, the estimate of the hardenability obtained with the tentative composition may be checked. Sample parts are then heat-treated and tested for hardening and for freedom from such metallurgical difficulties as temper brittleness. If incomplete hardening or other trouble occurs, changes have to be made in the composition, the heat-treatment, or the shape of the piece, and some of the steps of the design must be repeated. Finally, when the difficulties are ironed out, service tests can be made and process control procedures established.

Many of the fundamental metallurgical principles needed for design have not yet been developed, as will become apparent particularly after study of the chapters devoted to transformations, heat flow, and mechanical behavior. Many of the quantitative data required in the application of the known principles are not yet available. Metallurgical design is still in its infancy. In the following chapters, an attempt has been made to present and organize the currently known principles and data that are pertinent to the problem of metallurgical design of steel parts for optimum mechanical properties, in order to permit intelligent design and to point out those problems that require further study.

I. PHASE TRANSFORMATIONS IN STEEL

To produce a tempered-martensitic structure, it is necessary to heat-treat steel so that certain transformations do occur and certain others do not. To design the heat-treatment, it is necessary to have a thorough knowledge of the possible transformations, and of the conditions under which they take place. A discussion of this subject logically begins with a study of the directions toward which the transformations proceed, the so-called "equilibrium states". This topic lies in the field of thermodynamics. When the equilibrium conditions are understood, it is possible to consider the rates and the mechanisms of the transformations: topics in the field of reaction kinetics.

Equilibrium Conditions

Iron-Carbon Alloys

The iron-carbon phase diagram (Figure 1) is probably the most convenient starting point for a discussion of the transformations of steel. It is necessary to understand the meaning and limitations of this diagram, at least of that portion of it involved in hardening and tempering. This portion consists of four composition-temperature regions, separated by three boundary lines. Below the A_1 line the diagram indicates that the equilibrium (more exactly, quasi-equilibrium*) phases are ferrite and cementite. Between the A_1 and the A_3 lines the equilibrium phases are ferrite and austenite**. The A_3 line indicates the amount of carbon that austenite contains while in equilibrium with ferrite. This ferrite has the carbon content indicated by the line separating the field of ferrite from that of austenite plus ferrite. The A_{cm} line denotes the limit of solubility of carbon in austenite in equilibrium with cementite. Above the A_3 and

*Actually, the ordinary diagram, as given in Figure 1, deals with a metastable or quasi-equilibrium, rather than a stable equilibrium. The stable equilibrium diagram would show graphite rather than cementite throughout much of the diagram, and the extent of austenite field would be greatly reduced(2). However, both in ordinary steels and in iron-carbon alloys of corresponding carbon content, cementite ordinarily occurs rather than graphite, and its decomposition into the latter during hardening and tempering is generally so slow that it can be neglected. The phases of the metastable cementite diagram will, for simplicity, generally be referred to as equilibrium phases in the remainder of this volume, with the understanding that strictly they may be quasi-equilibrium.

**Ferrite is a phase having the lattice structure of body-centered-cubic iron. Austenite is a phase having the lattice structure of face-centered-cubic iron. Cementite is a phase having the orthorhombic lattice structure of iron carbide. A discussion of these structures can be found in the recent book "Structure of Metals"(3).

A_{cm} lines (and, over most of the carbon range, up to temperatures where melting begins) the equilibrium phase is austenite.

No further information (except a little concerning the relative quantities of each phase in equilibrium in the two-phase regions) can be obtained from the pertinent portion of the iron-carbon diagram. The diagram says nothing about the phases that will exist after any real heat-treatment, but merely describes the phases that would be approached if an iron-carbon alloy of any composition were held at constant temperature and atmospheric pressure for a sufficiently long time. Thus, consider an iron-carbon alloy containing .40 per cent carbon and originally consisting of ferrite and austenite, held at a temperature of 1800° F. (980° C.) and atmospheric pressure. The point corresponding to .40 per cent carbon, 1800° F. and atmospheric pressure is in the diagram field marked "austenite". The diagram, therefore, tells us that the steel will tend to consist of austenite alone; the amount of austenite will not decrease, and given sufficient time, it will increase, so that eventually the steel will be entirely austenitic. Whether a thousandth of a second or a thousand centuries will be required to produce an appreciable increase in the amount of austenite, the diagram does not say. Moreover, the diagram states nothing about the course of the reaction; whether or not, for example, ferrite that tends to decompose to austenite will first transform to some intermediate phase and the latter in turn transform to austenite. Nor does it give any information as to the shape, size, or distribution of phases, or as to their crystallographic orientation and habits. As drawn, the diagram does not apply if the pressure is other than one atmosphere, but the modifications produced by pressure changes are negligible unless the pressure changes are very great. Further, the diagram is for iron-carbon alloys, not for steels, even plain carbon steels, as steels contain many elements besides iron and carbon.

Effect of Additional Elements

When steel-making elements (such as manganese and silicon) or alloying elements are added to an iron-carbon alloy, the general features of the equilibrium diagram persist up to rather large percentages of these additional elements. However, the positions of the boundary lines are shifted; nickel and manganese, for example, move the A_3 line to the left. Another effect of the addition of elements is to resolve the boundary lines between two-phase regions into fields containing three or more phases. Thus, when nickel or manganese is added, the A_1 line resolves into a region in which ferrite, cementite, and austenite are all in equilibrium. (A phase diagram illustrating this three-phase region and the changes of the iron-carbon diagram effected by the addition of manganese is given as Figure 2.) Furthermore, there are differences in the significance of the diagram when additional elements are added; the iron-carbon binary diagram becomes a section (at constant percentage of the other elements) of a multi-constituent diagram, which thermodynamically is by no means the same thing. This alteration need not concern us here, but one point to be remembered is that new phases may be introduced which do not occur in the binary system. If, say, 4% chromium or 1-1/2% molybdenum is added to an iron-carbon alloy, new carbide phases, rich in these elements, may appear

in addition to cementite, and cause significant changes in the diagram. For example, the A_{cm} line* may be moved to the left (upward) so that austenite without carbides will be the equilibrium phase only at very high temperatures, if at all. The compositions of the ferrite, austenite, and cementite phases will also be altered by alloying elements.

The effects of each of the additional elements upon the iron-carbon diagram have not been fully investigated experimentally, though a moderate amount of data is available(5-8). Practically no experimental work has been done for alloys that contain two or more elements other than iron and carbon. Fortunately, by applying the principles of thermodynamics, it has been possible to fill in many of the gaps and to extend the knowledge to multi-component systems(2, 8). By combining experimental and theoretical results, it is found that the solubility of carbon in austenite in equilibrium with ferrite is decreased (A_3 shifted to the left) by such elements as nitrogen, manganese, nickel, zinc, and copper (in order of decreasing effect). Chromium and cobalt have little effect, while the boundary is shifted to the right, as shown schematically in Figure 3, by tungsten, molybdenum, silicon, vanadium, tin, aluminum, beryllium, phosphorus, and titanium (in order of increasing effect on a weight per cent basis). The shift is approximately proportional to the per cent of element added, and when several elements are present, their effects in changing the carbon content of austenite in equilibrium with ferrite are algebraically additive(2).

The effects of the various elements on the solubility of carbon in austenite in equilibrium with cementite (A_{cm} line) are somewhat more complicated. Small amounts of manganese, silicon, chromium, copper, molybdenum, and titanium (in order of increasing effect), and of tungsten and vanadium, move the A_{cm} line uniformly to the left. This decrease in solubility of carbon continues for high percentages of silicon, manganese, and copper, which have little tendency to form complex or alloy carbides. (Figure 4.) Nickel, for which no data exist, undoubtedly acts similarly.

The other elements, which do have greater tendency than iron to form carbides, introduce new carbide phases. When the amount of one of these so-called "carbide-forming" elements becomes sufficiently high, the phase in equilibrium with austenite immediately to the right of the austenite region becomes, at low temperatures, one of these other carbides instead of cementite. Under these conditions, austenite in equilibrium with the alloy carbides contains less carbon than austenite in equilibrium with cementite. The austenite region will thus be restricted at low temperatures to low carbon contents. At higher temperatures, the limiting phase will still be cementite and the high-carbon boundary of the austenite will not be shifted so far to the left. (Figure 4.) As the amount of chromium, tungsten, molybdenum, vanadium, or titanium increases, the bound-

*The term "A_{cm}" is used, for simplicity, to mean "the high-carbon boundary, below the melting temperature, of the region in which austenite is the equilibrium phase", even though the phase appearing immediately beyond this boundary is not cementite but some other carbide.

ary between austenite and austenite-plus-cementite is restricted to very high temperatures only and eventually disappears leaving in its place a boundary between austenite and austenite plus the alloy-carbide phase. This boundary will occur at very low carbon contents at low temperatures and at higher carbon contents at higher temperatures. The influence of temperature on the boundary will be greater than when cementite limits the austenite region. The entire boundary, however, will lie at lower carbon contents than for iron-carbon alloys.

The effects of several elements, present simultaneously, upon the A_{cm} boundary will superimpose, though they will not add linearly; quantitative descriptions of such cases, based on thermodynamic considerations, have been worked out(2).

Little information is available concerning the effects of alloying elements on the A_1 line. As mentioned above, in the presence of elements other than iron and carbon, this line resolves into a three-phase region, whose upper and lower boundaries are not necessarily horizontal. (Figure 2.) One point on the upper boundary of this three-phase region is fixed by the intersection of the A_3 line with the A_{cm} line. The intersection may be termed the "eutectoid point"; the temperature and carbon content at which it falls are called the "eutectoid temperature" and the "eutectoid carbon content"*. Elements such as tungsten, molybdenum, vanadium, titanium, and probably chromium, move the A_3 line to the right (or have little effect upon it) while moving the A_{cm} to the left. They thus raise the eutectoid temperature and hence raise part, at least, of the upper boundary of the three-phase region. The meager data available indicate that for such elements the lower boundary also is raised slightly. In the case of manganese, which moves A_3 to the left, both upper and lower boundaries of the three phase region are lowered, the shift being greater the lower the carbon content(4). Since little theoretical work has been done on the A_1 line, it is not possible to generalize these results.

Formation of Austenite

General

To harden steel, austenite must first be obtained. Since prior to the hardening operation the structure, resulting from a previous treatment, ordinarily consists of ferrite and carbide, the first reaction involved in hardening is the transformation of ferrite and carbide to austenite When a mixture of ferrite and carbide is heated into the temperature range where austenite is the stable phase, the ferrite and carbide react at their interface to form nuclei of austenite, which grow, absorbing the ferrite and the carbide(9-11). The process of nucleation and growth continues until the equilibrium percentages of the phases are present.

*Steels having lower carbon contents than the eutectoid are termed "hypoeutectoid" and steels having higher carbon contents, "hypereutectoid".

The nucleation and growth take time, and proceed more rapidly at high temperatures(10). The relations between time and temperature for austenitizing have not yet been established, but indications are that the reaction is speeded up tenfold by an increase of 50 to 100° F.*. The ferrite transforms to austenite more rapidly than the carbides dissolve, so that in a eutectoid steel undissolved carbides may remain when the ferrite has disappeared(10, 14-16). Alloy carbides are particularly slow to dissolve, and may persist even in hypoeutectoid steels long after the ferrite has transformed. High austenitizing temperatures are, therefore, necessary to obtain complete austenitization within a reasonable time in steels containing moderate or high percentages of the carbide-forming elements. (The most common of such elements are chromium, molybdenum, tungsten, vanadium, and titanium.) These elements, as explained above, tend to shift the A_{cm} line towards lower carbon contents and the A_3 towards higher carbon contents, thus restricting to high temperatures the range in which austenite can exist without ferrite or carbides under equilibrium conditions. These elements also slow the transformation of ferrite and carbide to austenite. This retardation is presumably connected with their effect in slowing the diffusion necessary for the transformation, but little quantitative information exists. Large percentages of carbide-forming elements** appear to slow the transformation to such an extent that it is not completed within times practical for heat treatment unless the temperature is high. It has been reported that deoxidizing additions of aluminum also decrease the rate of the austenitizing reaction(10). Whether or not this is simply caused by the alloying effect of residual aluminum has not been investigated.

The time necessary for austenitizing depends to some extent upon the prior structure. In general, the coarser the distribution of carbide and ferrite, the longer is the time necessary for complete austenitization. For a given steel and austenitizing temperature, a fine pearlite (fine alternate lamellae of ferrite and carbide) or a martensite (which tempers during heating to fine carbides distributed rather uniformly and finely in a matrix of ferrite) will transform completely to austenite sooner than a coarse pearlite with primary ferrite (coarse alternate lamellae of ferrite and carbide with regions of ferrite only) or a coarse spheroidized structure (coarse carbide spheroids spaced far apart in a matrix of ferrite)(10, 14, 17). The austenitic grain size of the prior structure does not seem to have any effect upon the time required for austenitizing(10).

Segregation

Even when the steel has wholly transformed to austenite (plus nonmetallic inclusions), concentration gradients may persist. Four types

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*Calculated for about 1650° F. from data of Roberts and Mehl(12) and of Williams(13).

**The percentage considered "large" will depend on the element. To give a rough idea, 2.00% chromium, 1.00% tungsten, .50% molybdenum, .25% vanadium, or .15% titanium might be considered a "large" amount.

of segregation may be distinguished on the basis of their scale:

1. Segregation resulting from concentration differences between carbide lamellae and neighboring ferrite lamellae, or between carbide spheroids and the neighboring ferrite matrix. (If several types of carbide are present, there may be concentration differences between them.) The spacing of the segregation in this case is of the order of 10^{-7} to 10^{-3} inch.

2. Segregation resulting from concentration differences between prior constituents, such as primary ferrite or primary cementite on one hand and pearlite, bainite, or martensite on the other. The spacing of such segregates is of the order of the prior austenitic grain size, ordinarily 10^{-3} to 10^{-2} inch.

3. Dendritic segregation arising during solidification of the steel. The spacing is of the order of about 10^{-2} to 10^{0} inch.

4. Large-scale segregation, also arising during solidification of the steel, such as that from the surface to the center of a casting or from the top to the bottom of an ingot. The scale is of the order of size of the casting or ingot, 10^{0} to 10^{+2} inch.

These concentration differences persist to a certain degree when the steel has transformed to austenite, and time is required to permit the diffusion necessary to eliminate them. The rate of their elimination or reduction increases rapidly with the temperature in the same way as do the diffusion coefficients; the reciprocal of the rate varies exponentially as the reciprocal of the absolute temperature. Since the diffusion coefficient of carbon is much greater than that of the ordinary alloying elements (except nitrogen, if nitrogen is considered an alloying element) carbon segregation tends to equalize more quickly than alloy segregation*. Calculations based upon known(18) diffusion coefficients indicate that carbon segregation resulting from concentration differences between carbide lamellae or spheroids and the adjacent ferrite should disappear in times of the order of a second or less after austenitization is complete. Carbon differences from boundary to center of prior austenite grains are

*A question might be asked as to the effect of alloying elements upon the diffusion of carbon. The effect of nickel, manganese, and molybdenum upon the diffusion coefficient of carbon in austenite has been investigated and found to be small(18, 19), and it might be concluded that the effect of the other alloying elements is also small. It has been suggested, on the basis of studies of the impregnation of steel with various elements(20-25), that these elements "repel" or "attract" carbon, so that segregation of these other elements might cause segregation of carbon to arise or persist in austenite. While thermodynamically such an effect is possible, it would be expected to be small, and it appears that none of the studies mentioned dealt with austenite in the absence of other phases. They are, therefore, not applicable to the case under consideration.

smoothed out in times ranging from less than a second to several minutes, depending on the austenitizing temperature and the austenitic grain size of the prior structure. (Austenitizing temperatures of 1330° F. (720° C.) to 1830° F. (1000° C.) are considered here.) Fine dendritic segregations of carbon persist only a few seconds at high austenitizing temperatures and a few minutes at low, but coarse dendritic segregates may last days at high and weeks at low austenitizing temperatures*. Large-scale carbon segregation is not appreciably affected for days or years at ordinary austenitizing temperatures.

For alloying elements the rates of homogenization are much slower and depend to a far greater extent upon temperature**. Alloy segregation associated with very fine lamellae or spheroids disappears in a fraction of a second at any austenitizing temperature. Alloy segregates resulting from very coarse lamellae or spheroids last a few minutes in low-carbon steel at high austenitizing temperatures and perhaps weeks at low temperatures. In high-carbon steels (1.00% carbon or more) they can persist for hours at high and years at low temperatures***.

Alloy segregates at the grain boundaries of very fine prior austenitic grains can be removed in some hours at a high austenitizing temperature, while with coarse prior grains or a low temperature, weeks or years will be required. Interdendritic and large-scale segregation of alloying elements will persist for weeks or years at any normal austenitizing temperature.

The presence of concentration gradients of carbon or alloying elements causes differences in the temperature at which the austenite can decompose. Further, their presence will, as discussed later, cause the decomposition of the austenite to occur at different rates in different regions of the metal.

Austenitic Grain Size

When the austenite grains forming from ferrite and carbide have

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*The high and low austenitizing temperatures mentioned here and in the following paragraphs refer to the 1330° F. and 1830° F. in the preceding sentence.

**The times given for alloying elements are based upon recent studies of the diffusion coefficients of manganese(26), nickel(27), and molybdenum (28) in austenite. The coefficients are similar for the three elements.

***The times for homogenization of concentration differences originating as differences between carbide lamellae or spheroids and the adjacent ferrite are greater in high-carbon than in low-carbon steels because the thickness of the carbide lamellae or spheroids increases with the carbon content for constant spacing. The spacing does not decrease much with increasing carbon content for constant temperature of formation or degree of tempering(29, 30).

grown to such an extent that their boundaries touch, further growth of each grain is limited by the presence of neighboring austenite grains. After the austenite grains contact each other, some may grow at the expense of their neighbors, absorbing them. The size of the grains then depends both upon the initial size established when the austenite grains first touch, and upon the extent to which they grow thereafter. From the data available(10, 31), it appears that the initial grain size may either increase or decrease as the rate of heating is increased, depending, presumably, upon the particular steel, prior structure, and range of heating rate employed. It may be deduced that the effect of austenitizing temperature upon the initial grain size is similarly indeterminate. (If the heating is not very rapid, it is possible for the initial grain size to be established before the steel reaches the selected austenitizing temperature.) Prior structure can have some effect; it has been shown that fine lamellar structures result in a slightly finer initial austenitic grain size than coarse lamellar structures(10) and it would appear that the same would be true for fine and coarse spheroidal structures. The effect of composition upon the initial grain size apparently has not been investigated, except that suitable small additions of aluminum have been shown to decrease it(10).

After the grains have contacted each other, they will ordinarily grow at each other's expense, at least if the temperature is high enough. The time-temperature relationships have not been fully investigated, but it appears that a considerable change in time is required to produce an effect equivalent to a small change in temperature. It is frequently stated that some minimum temperature is required with all steels for grain growth to occur, but the evidence of this is hardly conclusive(32). The temperature at which grain growth occurs rapidly can be raised considerably by suitable additions of aluminum. Recent experiments indicate that this effect is associated with the presence of aluminum nitride(33), presumably in the form of fine (probably submicroscopic) inclusions. (Earlier investigators supposed that aluminum oxides prevented the growth.) The effect is found only if the amount and time of the aluminum addition is suitably adjusted to the rest of the melting practice; this appears to be a matter of producing the proper type, size, and number of inclusions. The retardation of austenite grain growth produced by these aluminum (nitride) inclusions is simply one example of the general phenomenon, not very well understood, that the grain growth of a phase may be markedly retarded by the presence of even a very small amount of another phase, suitably distributed(32). Another example is the retardation of austenite grain growth in vanadium-containing steels by undissolved vanadium carbide particles; rapid growth may occur when the temperature and time are sufficient to dissolve these carbides(34). In general, little grain growth occurs in steel containing undissolved ferrite or carbide. The effect of alloying elements in retarding austenite grain growth(35) may be due simply to the undissolved carbides, which are more likely when alloying elements are present. Besides aluminum and vanadium, titanium and zirconium are sometimes added to steel for the purpose of retarding grain growth(36). The composition of the undissolved particles which they may be supposed to form has not been determined; oxide(36), nitride(36), and carbide(34) have been suggested.

Coarsening in steels with growth-retarding additions has been studied primarily in cases where the addition was aluminum. For these steels, as the temperature is increased, no appreciable coarsening takes place in an initially fine-grained austenite until a particular temperature (referred to as the "coarsening temperature") is reached. Above this temperature a few of the grains grow markedly, while the others do not grow significantly. As the temperature is further increased, more of the grains grow abruptly, until all of the small grains are absorbed(32, 35, 36). On further heating, more gradual growth occurs, similar to that ordinarily found in steels to which no growth-retarding additions were made. This behavior might indicate that at the coarsening temperature the aluminum nitride inclusions go into solution, removing the restriction to grain growth, but there is no experimental evidence to support this hypothesis.

The term "grain size" when used in connection with steel has come to mean the austenitic grain size. When the structure does not consist of contiguous austenite grains, "grain size" refers to the prior austenitic grain size; the austenitic grain size existing when the structure did last consist of contiguous austenite grains. "Grain size" does not refer to the size of the ferrite or carbide grains that may exist in the structure under study. Grain size is most commonly measured by the A.S.T.M. scale, which is expressed in terms of the average size of the intersections of the grains with a plane, and not directly in terms of the size of the three-dimensional grains. The A.S.T.M. grain size is defined by the equation,

$$a = 2^{n-1}$$

where n is the grain size and a is the number of grains per square inch as viewed at 100 diameters magnification. In practice, the grain size is usually determined by comparison with standard charts or samples(37). When some of the grains are very much larger than others (so-called "mixed grain size"*, as may occur when some grains have grown considerably but have not absorbed all the initial small grains), it is customary to rate the large and the small grains separately, rather than to average them together. Commercially, A.S.T.M. 4 and coarser sizes are considered "coarse", while A.S.T.M. 5 and finer are considered "fine".

Methods for revealing the prior austenitic grain size at room temperature have been established(37). Some of these methods involve carburizing or oxidizing the steel and are not as desirable as the others, because changes in carbon or oxygen content may affect the grain size. Austenitic grain size is not affected by the rate of cooling from the austenitizing temperature or by heat treatments at temperatures below the A_1 range. The size established by a known austenitizing treatment may, therefore, be determined by methods which involve cooling from this treatment at any rate that produces a structure delineating the grains. This rate

*The term "duplex grain" is often used instead of "mixed grain size".

may be different from that to be employed in obtaining the desired structure. One general method of determining the grain size established in a steel by an austenitizing treatment is to cool after the treatment at such a rate that one constituent (ferrite, carbide, pearlite) precipitates at the grain boundaries while another forms at the centers of the grains, and then examine the resulting structure under the microscope. Another metallographic method involves the use of an etching reagent that develops contrast between the martensite formed from different austenite grains. This is applicable to structures that consist of martensite (with or without retained austenite) at room temperature and to austenite that can be converted to such structures by quenching. A third method, applicable to steel at high hardness levels, consists of fracturing a piece with little deformation and comparing the appearance of the fractured surface with a set of fracture standards. This method is probably not very reliable for hypoeutectoid steels. Details of the various methods can be found in the literature(37).

Steels in which austenitic grain growth is restrained are commonly referred to as "inherently fine-grained" steels. Steels for which the rate of growth increases uniformly with temperature are referred to as "inherently coarse-grained" steels. These terms, therefore, bear no relation to the austenitic grain size actually existing in the steel, which may be coarse or fine, depending upon the treatment given, regardless of whether the steel is inherently coarse- or fine-grained. There is some evidence that the coarsening characteristics do not depend solely upon the composition and melting practice, but can be affected by high-temperature treatments in the solid state(38).

Decomposition of Austenite

Isothermal Decomposition

If austenite is cooled below its range of stability, it will eventually decompose. The manner, progress, and product of the decomposition depend upon both temperature and time. Perhaps the simplest method of studying the process is to consider the transformation of austenite cooled very rapidly from within its stability range to some lower temperature and then held at that temperature. This approach to the problem has the advantage that, if the initial cooling is rapid enough so that no reaction takes place until the holding temperature is reached, the process occurs essentially at a constant temperature (isothermally), and its progress is not complicated by the effects of changes in temperature.

Figure 5 illustrates the times at which isothermal decomposition begins and ends in a steel containing a moderate percentage of carbide-forming elements. Figure 6 is a schematic diagram constructed by principles to be discussed later. There are two ranges of rapid isothermal decomposition. The first of these ranges occurs at about 1150-1250° F. (620-675° C.), and is associated with the decomposition of austenite into lamellar pearlite and proeutectoid ferrite or cementite. The second, at 750-950° F. (400-500° C.), is related to the decomposition of austenite into bainite.

The third range indicated in Figure 5 is associated with the composition of austenite into martensite and is found only upon cooling (not isothermally). The temperature of transformation of austenite to martensite is greatly influenced by composition, and varies from approximately 1000° F. (550° C.) to below room temperature.

In steels not containing carbide-forming elements, such as plain-carbon steels, only a single range of rapid isothermal decomposition has been generally reported (Figure 7). This failure to find separate pearlite and bainite "noses", separated by a range of slow transformation, may be due to the experimental difficulties involved in accurate isothermal work at very short times, for continuous cooling measurements(41, 42) seem to indicate that there are two noses in plain-carbon (low-manganese) steels. Moreover, recent work(43, 44) shows that overlapping of the pearlite and bainite reactions occurs so that there need not be a range of slow transformation between the temperature ranges associated with each.

Pearlite and Proeutectoid Products

If homogeneous austenite is cooled to a temperature range where austenite and ferrite or austenite and carbide are the equilibrium phases, and held in this range, the ferrite or carbide precipitate. This process takes place by nucleation and growth, and requires time. Nuclei appear at the boundaries of the austenite grains, and growth takes place in such a way that the proeutectoid constituent (ferrite or cementite) first appears as small patches along the grain boundaries. These patches grow until a continuous layer is formed at the grain boundaries, provided equilibrium considerations permit sufficient proeutectoid product to precipitate. Further growth merely thickens the grain-boundary layer. Figure 8 includes photomicrographs showing grain boundary precipitates of ferrite and of cementite. If, at the temperature of transformation, ferrite is the equilibrium phase, the reaction will continue until all the austenite has transformed; this occurs only when the carbon content is very small. When, as is usual, the equilibrium state requires the presence of some austenite, this austenite will remain untransformed at the center of the grains. The equilibrium diagram indicates that, for a given steel, the amount of the austenite which remains will be smaller the lower the temperature.

Since the primary ferrite that forms is lower in carbon than is the austenite, diffusion of carbon from ferrite to austenite takes place during the austenite to ferrite reaction, and the carbon content of the austenite increases as this reaction proceeds. Since the primary carbide that forms is higher in carbon than the austenite, diffusion of carbon from austenite to carbide takes place during the austenite to cementite reaction. The carbon content of the austenite decreases as this reaction proceeds. Diffusion of alloying elements so as to approach the equilibrium alloy compositions of the various phases must also take place, but whether the rate at which this occurs is comparable to the rate at which the precipitation occurs, or much slower, has not been determined.

Isothermal precipitation of ferrite from austenite is accompanied

by an expansion in volume. This expansion is of the order of one per cent, but increases considerably with increasing carbon content of the austenite and with decreasing temperature. Isothermal precipitation of cementite from austenite is accompanied by a slight contraction.

In discussing the decomposition of austenite below the A_1 temperature, it is helpful to refer to the quasi-equilibrium diagram for iron-carbon alloys as presented in Figure 9. This diagram illustrates the stability relations of austenite, ferrite, and cementite with respect to each other, below, as well as above, the A_1 temperature. The A_3 and A_{cm} lines extended below A_1 give the carbon content of the austenite that will be in equilibrium with ferrite, and of the austenite that will be in equilibrium with cementite, if the austenite is supercooled below the A_1 temperature. The extensions of the A_3 and A_{cm} lines have as much physical meaning as the phase boundaries themselves(46). The significance of the lines labelled "upper limit of bainite formation" and "upper limit of martensite formation" will be discussed later.

If austenite is cooled rapidly into a temperature range where both ferrite and cementite are stable phases, and held in this range, both ferrite and cementite will eventually form. Let us consider now only that part of this temperature region commonly referred to as the "pearlite range", which is ordinarily above 1000° F. (550° C.). The first reaction to take place here may or may not be the primary formation of ferrite (in hypoeutectoid steels) or carbide (in hypereutectoid steels). The likelihood of these reactions is greater the higher the temperature and the further the composition deviates from that of the eutectoid. Quantitatively, it appears that primary ferrite will form if the temperature of holding is below the A_3 line, but not below the extended A_{cm} line (Figure 9). For iron-carbon alloys, the diagram indicates that this will always be the case when the carbon content is below .40%. (If alloys are present that move the A_{cm} line to the left, ferrite may not form even with less than .40% carbon.) The primary precipitation of ferrite will proceed, in the manner described above, until the carbon content of the remaining austenite has increased to that indicated by the extended A_{cm} line. (Because of the diffusion gradients associated with the precipitation of the primary constituent, the carbon content of the austenite will not be uniform and some regions will reach this concentration before others.) Similarly, primary formation of carbide will take place if the temperature of holding is below the A_{cm} line but not below the extended A_3 line(40). For iron-carbon alloys, at least, the diagram (Figure 9) indicates that this need not occur below 1200° F. (650° C.). When it does occur, precipitation of carbide will proceed, in the manner described above, until the carbon content of the remaining austenite has decreased to that indicated by the extended A_3 line.

When the austenite is held at a temperature below both the extended A_3 and extended A_{cm} lines or when sufficient primary precipitation has taken place to bring it below them, simultaneous precipitation of fer-

rite and carbide takes place in the form known as pearlite*. This constituent consists of alternate plates or lamellae of ferrite and carbide, as illustrated in Figure 8. It forms by a process of nucleation and growth, the nucleation occurring at the austenitic grain boundaries(45). The formation of lamellae is associated with the necessity for the diffusion of carbon through the austenite from areas adjacent to newly-formed ferrite to areas adjacent to newly-formed carbide (Figure 10)(46). The distance between these areas is a minimum, if ferrite and carbide grow into the austenite at the same speed as alternate plates growing edgewise. The rate of growth, governed by the diffusion of carbon, is then a maximum, and the lamellar habit, therefore, soon predominates over any other habit with which the reaction might begin.

The relationships governing the rate of growth of the pearlite lamellae are such that this rate first increases and then decreases as the temperature is lowered**. The rate of nucleation increases continuously down to the lowest temperatures of the pearlite range. The effect of these variations of the rates of growth and nucleation with temperature is that the pearlite reaction is slow at the highest temperatures at which it occurs, more rapid at lower temperatures, and slow again at still lower temperatures(40). Several studies of the kinetics of pearlite formation, as governed by the rates of nucleation and growth, have been made(12, 47-61). The temperature range at which the transformation is most rapid, often called the "pearlite nose" of the isothermal transformation diagram, is commonly between 1150 and 1250° F. (625 and 675° C.). The pearlite transformation does not cease to occur below 1000° F. (550° C.), but it is slow there and ordinarily transformation to bainite occurs before the pearlite reaction can start.

The primary precipitation of carbide commonly occurs only at temperatures above the pearlite nose, and proceeds more rapidly the lower the temperature. In a few chromium-bearing steels primary precipitation of carbide occurs also at temperatures below the pearlite nose (in the form of thin plates along certain crystallographic planes rather than as a

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*It is possible for primary precipitation of ferrite, carbide, or both to continue to some extent in the region mentioned. This may perhaps be what occurs in so-called "abnormal" steels(36).

**The rate of pearlite growth varies directly as the rate of diffusion of carbon through the austenite. This rate decreases as the diffusion coefficient of carbon in austenite decreases and as the difference between the carbon concentrations of austenite in equilibrium with ferrite and with carbide decrease. It increases with decreasing spacing of the lamellae. The reciprocal of the coefficient of diffusion, in turn, varies exponentially as the reciprocal of the absolute temperature, so the diffusion coefficient decreases when the temperature decreases. The difference between the carbon concentrations is the difference between the extended A_3 and A_{cm} lines, and, as Figure 9 indicates, varies nearly directly as the difference between the temperature of transformation and the A_1 temperature. The spacing of the lamellae varies inversely as this temperature difference, because of surface and volume free energy conditions.

grain-boundary layer)(62, 63). In this temperature range the precipitation is slower the lower the temperature. However, the time for initiation of carbide precipitation changes less with temperature than the time for initiation of the pearlite reaction.

The primary precipitation of ferrite may, in many steels, occur both above and below the pearlite nose. Down to the pearlite nose, it is more rapid the lower the temperature. Whether further decrease of temperature, below the pearlite nose, delays or hastens the primary ferrite reaction has not been settled; possibly the effect may be different in different steels(39, 44).

The transformation of austenite to pearlite is associated with a change in volume corresponding to the changes involved in precipitation of ferrite and carbide from austenite. When the carbide is cementite, its precipitation is accompanied by little volume change. The expansion associated with pearlite formation is then of the same order as for the precipitation of ferrite from austenite, and increases as the transformation temperature is lowered.

If the steel is held sufficiently long at the transformation temperature, the pearlite lamellae will tend to break up, forming carbide spheroids* in a matrix of ferrite. The layers of proeutectoid ferrite or carbide will also reshape, the structure tending towards carbide spheroids in a ferrite matrix. These are simply examples of tempering phenomena, which will be discussed later. A case not often considered as an aspect of tempering occurs when decomposition of austenite takes place at a temperature above the A_1 range, so that one proeutectoid constituent forms at the grain boundaries but the centers of the grains remain austenitic. If the steel is held sufficiently long at this temperature, the layers of grain boundary ferrite or carbide will reshape into spheroids, in a matrix of austenite.

As the proeutectoid and pearlite reactions are nucleated primarily at austenitic grain boundaries, they take place more quickly the greater the grain boundary area per unit volume; that is, the finer the austenitic grain size(45, 64).

Alloying elements in the quantities ordinarily used in heat-treatable steels do not change the mechanism of the proeutectoid and pearlite reactions. However, almost all the elements so far investigated appear to retard the reactions. These elements include aluminum, antimony, arsenic, beryllium, boron, carbon, copper, manganese, molybdenum, nickel, phosphorus, silicon, tin, tungsten, vanadium, and zirconium(65-68). Cobalt is an exception, hastening the reaction, and one report indicates that germanium has

*It is often stated that spheroids of carbide may form directly from homogeneous austenite. This type of decomposition is extremely unlikely and may only occur(70) when the austenite from which they form is not homogeneous.

no effect*(68). There have been few systematic quantitative studies of the effects of any of the elements. It has been established(69), however, that molybdenum additions of less than 1 per cent can delay the apparent start and end of the pearlite and ferrite reactions by a factor of 1000. The effect of manganese is of the same order of magnitude, while that of chromium is somewhat less and those of nickel, silicon, and copper still smaller(65, 68). Carbon in small percentages has a very great influence upon the rate of these reactions, which proceed very rapidly when the carbon content is very low. At high percentages (over, say, .40%), the influence of carbon is not so great, being less than that of corresponding amounts of chromium. When several elements are present, their retarding effects upon the pearlite and proeutectoid reactions appear to reinforce each other, so that the times required for the reactions to become detectable and for the reaction to go to completion are greater than if any one of the elements were present alone(65). More quantitative data for the effects of the various elements, singly and in combination, upon the times required for the pearlite and proeutectoid reactions are available for continuous cooling than for isothermal conditions. The data for the case of continuous cooling and their relation and application to isothermal transformation will be discussed later.

When the structure before transformation is not austenite but a mixture of austenite with other phases (carbide, ferrite, sulfide, etc.,), the composition that primarily determines the rate of reaction is the composition of the austenite, rather than the average composition of the steel. Thus, suppose large quantities of molybdenum, for example, are added to a steel, and the austenitizing temperature and time are insufficient to dissolve the molybdenum carbides. The austenite then contains considerably less carbon (and only slightly more molybdenum) than the original molybdenum-free steel, provided the two steels are given the same austenitizing treatment. If both steels are rapidly cooled, following austenitizing, to a temperature within the pearlite range and held there, the steel with molybdenum will transform to pearlite more rapidly than the steel without molybdenum, simply because its austenite is of lower carbon content. The undissolved particles of another phase may also act as nuclei or points of nucleation for the formation of pearlite or proeutectoid constituents, and so further hasten their appearance. This effect is more pronounced at high temperatures, where nucleation in the absence of undissolved constituents is slow, than at low temperatures, where nucleation is relatively rapid in any case(12). The nucleating effect of undissolved particles, particularly at high temperatures, may change to some extent the habits of the constituents(70).

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*Some data are available which might be misinterpreted as indicating that columbium(68), sulphur(65), tellurium(65), and titanium(65), as well as large amounts of vanadium(65, 67), hasten the reaction. However, it appears that these data were not based upon an original structure of austenite alone, but upon structures containing also undissolved compounds of columbium, titanium, or vanadium with carbon (and possibly with nitrogen) or of sulphur or tellurium with manganese.

The persistence of inhomogeneities in austenite, even after the solution of other phases is complete, has been discussed above. At temperatures high in the pearlite range such inhomogeneities accelerate the initial appearance of pearlite. If the segregation is on a large or moderate scale, the regions low in alloying elements and in carbon react quickly but the regions high in alloying elements and in carbon might be expected to react slowly, increasing the time for completion of the transformation(71). For segregations on a small scale, it is possible that some of the small inhomogeneities speed the formation of correspondingly small areas of pearlite (or of ferrite or carbide) which in turn nucleate the transformation of the remainder of the austenite. In this latter case the completion of the transformation may not be delayed. That the second type of behavior holds for carbon segregation resulting from fine cementite spheroids is indicated by some evidence that this type of segregation has greater effect upon the reaction at high temperatures than at low(12), just as do undissolved constituents.

Variations in austenitizing treatment, then, have two primary effects upon subsequent transformation of the austenite in the pearlite range. One arises through the variations of austenitic grain size, the other through variations in solution of other phases in austenite and in the homogeneity of the austenite.

Bainite

If austenite is rapidly cooled to a temperature below about 1000° F. (550° C.) but above the temperature at which martensite forms, and held there, it will eventually start to transform to bainite. This constituent consists of ferrite and carbide, but there has been much confusion as to how these products form. It has even been questioned(51) whether the bainite reaction proceeds by nucleation and growth, but there appears to be sufficient evidence that it does.

Recent studies indicate that, in the bainite reaction, the austenite first transforms to ferrite (supersaturated) without diffusion of carbon but by a change (diffusion) of the lattice positions of the iron atoms(72, 73). It has been shown(40) that if austenite of a given carbon content is cooled sufficiently, it is unstable with respect to ferrite of the same carbon content. The temperature below which austenite will transform to supersaturated ferrite depends upon alloy and carbon content in about the same way as the A_3 line. (Figure 9.) The ferrite that forms is unstable and tends to decrease in carbon content. The carbon may diffuse into the adjacent austenite, or may simply diffuse to the boundaries of the ferrite and there precipitate. The carbon at low temperatures will precipitate within the ferrite areas. Thus, the final product of the bainite reaction is a mixture of ferrite and carbide.

If the adjacent austenite becomes so enriched in carbon that it will no longer transform to supersaturated ferrite, the austenite can no longer decompose in this fashion. (For iron-carbon alloys the calculated composition of the stable austenite is marked "upper limit of bainite formation" in Figure 9.) Measurements of the martensite-start temperature of

austenite remaining after cessation of the bainite reaction indicate that this austenite is in fact of higher carbon content than the original austenite(44, 63, 74-78). Further isothermal transformation of this austenite can be only to some product other than bainite.

The lower the temperature, the less likely is diffusion of carbon into the surrounding austenite, and the higher must be the carbon content of the austenite before the transformation will cease. Thus, the lower the temperature the greater is the extent of the transformation to bainite and the less austenite will remain, unable to transform(43, 44, 76-78).

The supersaturated ferrite appears to form in plates, parallel to certain crystallographic planes of the austenite (not very different from the (111) planes). The carbides, at least at high temperatures, seem to precipitate in plates parallel to the ferrite plates. These carbide plates spheroidize if held at the temperature of formation. At low temperatures the carbides appear to precipitate as small particles within the ferrite. Photomicrographs illustrating typical bainite structures are included in Figure 11.

Isothermal studies indicate that the rate of the bainite reaction is substantially independent of grain size(43, 64), and this might seem to indicate that the reaction does not occur preferentially at grain boundaries. However, microstructural studies of partially transformed steels show more bainite near grain boundaries than near the centers of grains(69, 79). Thus, while most of the bainite may be nucleated within the grains, there is some tendency towards preferential nucleation at the boundaries.

The time necessary for the bainite reaction to become detectable increases with decreasing temperature over most of the bainite range (Figures 5 and 6). Although, as mentioned above, the bainite reaction apparently will not occur above a certain temperature for each steel, there seems to be no minimum temperature. If the temperature is low enough for martensite to form, it does so first, but austenite not transformed to martensite may transform isothermally to bainite. Formation of bainite, because of the lower temperatures, is accompanied by an expansion somewhat greater than that associated with the formation of pearlite.

Alloying elements in the quantities ordinarily used in heat-treatable steels do not alter the mechanism of the bainite reaction. Their diffusion coefficients at the temperatures involved are so low that appreciable diffusion of alloying elements during the bainite transformation seems impossible. Nevertheless, alloying elements have appreciable effects upon the rate of the transformation, presumably through their effect upon the equilibrium relations among the various phases concerned. Those elements that shift the A_3 boundary to lower carbon contents should also shift the "upper temperature of bainite formation" (Figure 9) to the left, for the same reasons. The bainite reaction should then occur at lower temperatures and hence at slower rates(40). The elements that act in this way have been listed earlier. Though there are some inconsistencies, the results of this reasoning are very nearly in accord with the measured effects of alloying

elements. Of the elements studied, carbon and manganese seem to have the greatest effects, delaying the bainite reaction in about the same proportion as they do the pearlite(39, 71). (Carbon may even have a slightly greater effect upon the bainite than upon the pearlite transformation.) The effect of chromium is not well known. It appears to be appreciable, but not as great as the effect chromium has upon the pearlite reaction(44). Indications are that nickel(39) and probably silicon(80) have small effects on the bainite transformation, delaying it in about the same proportion as they do the pearlite. Molybdenum(69), on the other hand, which has a very great effect upon the rate of pearlite formation, has little if any effect upon the isothermal formation of bainite, as illustrated in Figure 12. Thus, as far as present rather fragmentary studies go, it appears that non-carbide-forming elements delay the bainite reaction in about the same proportion as they delay the pearlite, while carbide-forming elements have less effect upon the bainite than upon the pearlite reaction.

As in the case of the other types of decomposition to which austenite is subject, it is primarily the composition of the austenite, rather than the overall composition of the steel, that determines the rate of transformation. The nature of the effects of undissolved carbides and ferrite and of segregation in the austenite are probably similar for the bainite and the pearlite reactions.

Martensite

The martensite reaction, which takes place at temperatures lower than the pearlite reaction and ordinarily lower than the bainite, cannot well be considered isothermally because it is not a function of time as are the pearlite and bainite transformations. The start of the martensite reaction seems to occur at a definite temperature for each composition of austenite, and austenite apparently cannot be quenched wholly undecomposed to any lower temperature regardless of the speed of quenching(41, 81-86). Details of the martensite reaction can be profitably studied only on continuous cooling.

Relations Among Stages

If a steel is permitted to transform isothermally at a temperature at which first bainite and later pearlite forms, it is found that pearlite appears earlier than the time-temperature plot for its appearance at the higher temperatures (where no bainite forms) would lead one to expect(44, 63). Thus, bainite apparently nucleates the transformation of the remaining austenite to pearlite. The experiment described above has been performed only for temperatures close to the upper limit of the bainite range, where little bainite forms. At somewhat lower temperatures, where more bainite forms, it is possible that the increase in carbon content of the remaining austenite is sufficient to delay the pearlite transformation beyond the time at which it would be expected if no bainite were present.

Consider next specimens cooled sufficiently so that there is some transformation to martensite, and then held at this temperature so that the

remainder will start to transform isothermally to bainite. It is found that bainite appears earlier than the time-temperature plot for its appearance at higher temperatures (where no martensite forms) would lead one to expect(87-89). Thus martensite apparently nucleates the transformation of the remaining austenite to bainite. Effects of holding in the pearlite and bainite ranges upon later transformation to martensite will be reviewed after the nature of the martensite transformation has been discussed.

A more general question arises when austenite is cooled rapidly to a temperature below the critical range and held there for such a time that it either does not transform or transforms only in part, and subsequently cooled rapidly to a lower temperature and held there. The question is, what is the quantitative effect of the holding at the first temperature upon the rate of transformation at the second temperature. Consider first the case where at both temperatures the only product expected is pearlite. Several investigators have suggested that the fraction of the austenite transformed at each temperature is the important quantity(90), or if it is convenient to consider that a finite time is required for the transformation to begin, the fraction of this time elapsed(48, 59, 91-93). (These two versions of the hypothesis are similar in concept and more or less complementary.) For example, according to the first version, if the austenite is held 10 minutes at 1200° F., during which 1% of the austenite transforms, and is subsequently quickly cooled to 1100° F., the reaction will continue at 1100° just as if 1% of the austenite had transformed at 1100°. Thus, if a specimen quenched directly to 1100° F. from the austenitizing temperature would require 5 minutes for 1% transformation and 7 minutes for 2%, the specimen transformed 1% at 1200° will only need 2 minutes at 1100° for 2% transformation. According to the second version of the hypothesis, if the austenite is held 1 minute at 1200° F., where the reaction starts only after 5 minutes, and the austenite is then quickly cooled to 1100° and kept there, it can be considered that the austenite has been held for 1/5 the time necessary for the reaction to begin. If this time at 1100° is 3 minutes, then it can be considered that the austenite has been held 3/5 minute, and the transformation will begin in an additional 2-2/5 minutes.

The second version of the hypothesis, dealing with the time for transformation to begin, has been verified experimentally for one steel(93). The first version, dealing with fraction transformed, has been checked only indirectly(90).

Although similar hypotheses have been advanced for the cases where, at both sub-critical temperatures, ferrite, cementite or bainite(90) is the first decomposition product to appear, there is as yet no experimental evidence to support them. Neither has any information been published as to the effect upon subsequent transformation to bainite of holding at a temperature where the first product to appear is pearlite, except in high speed steels where some data(78) indicate that holding in the pearlite region accelerates bainite transformation.

When austenite is held at a temperature where the first decomposition product to appear is ferrite, subsequent transformation to bainite

(at a lower temperature) is accelerated(79). It appears that traces of ferrite, formed (along the grain boundaries) at the higher temperature, nucleate the formation of bainite. On the contrary, when austenite is held at a temperature where the first decomposition product to appear is cementite, subsequent transformation to bainite (at a lower temperature) is retarded(79, 94). Presumably cementite forms at those dislocations and irregularities where bainite would otherwise tend to nucleate most readily. By "using up" these discontinuities, cementite delays subsequent nucleation of bainite. It may be noted that the times reported for the first appearance of bainite, isothermally, are much greater in hypereutectoid steels than in hypoeutectoid steels of the same alloy content. The difference is considerably greater than would be expected for the increase in carbon content. This difference may be due to the influence upon the bainite reaction of slight traces of cementite and ferrite, respectively. These traces may form during the cooling from the austenitizing temperatures to the temperature of isothermal transformation.

Decomposition on Continuous Cooling

As mentioned earlier, the formation of martensite occurs only upon continuous cooling. Steels in which this structure is desired must, therefore, undergo transformation during continuous cooling, and the pearlite, bainite, and proeutectoid reactions must be avoided during this cooling.

Proeutectoid, Pearlite, and Bainite Reactions

It appears that transformation to proeutectoid ferrite and carbide, pearlite, and bainite during continuous cooling can be considered as isothermal transformation at a series of successively lower temperatures. For these transformations the steel can be considered to be held at each temperature for a short time and then cooled very rapidly to the next lower temperature. (During the very rapid cooling, no transformation is assumed to take place.) If the temperatures are equally spaced, the time the steel is considered held at a temperature will be inversely proportional to the actual cooling rate at that temperature. On the basis of assumptions mentioned in the preceding pages, it is then possible to calculate the decomposition behavior of austenite during continuous cooling, if its isothermal behavior is known(59, 90-93, 95). Such calculations have been checked experimentally, in a quantitative fashion, for only one case, that of pearlite formation in a plain-carbon steel containing 1.20% carbon(90). However, qualitatively, the decomposition on continuous cooling that is calculated as suggested accords with that found experimentally. Slow cooling rates mean, essentially, that the assumed time at each temperature is long, while fast cooling rates mean that the time at each temperature is short. An example of corresponding diagrams for the beginning of decomposition to pearlite, isothermally and on continuous cooling, is presented as Figure 13.

On this basis, if the cooling rate through the temperatures of the pearlite and proeutectoid reactions is very slow relative to the time required for these reactions, they will occur at high temperatures and will

continue until all the austenite has decomposed. If the cooling rate is somewhat faster (relative to the time required for the pearlite and proeutectoid reactions), they may still occur and go to completion, but the reactions will take place at a lower temperature. The characteristics of the products formed, such as the interlamellar spacing of the pearlite, will be closely similar to those of products formed isothermally at the same temperature. However, since the decomposition of the austenite takes time, it does not, on continuous cooling, occur at one temperature, but over a range, and the characteristics of the products will show a corresponding range.

Unless the cooling rate decreases greatly as the temperature falls, little pearlite will form, on continuous cooling, at temperatures much below the pearlite nose. The times required for transformation at such temperatures are relatively great. As a consequence, appreciable transformation can occur at these temperatures only with slow cooling. With slow cooling, however, all the austenite will generally decompose at the nose temperature, where the times required are short, or before reaching the nose temperature. (Figure 13.)

If the cooling rate is sufficiently fast, the pearlite reaction will not go to completion, and some austenite will remain down to temperatures at which the bainite reaction takes place. At a still faster rate, the pearlite and proeutectoid reactions will not even start. In either of these last two cases, if the cooling is not too fast relative to the time required for the start of the bainite reaction, some of the austenite will transform to bainite. The characteristics of the bainite, again, will be similar to those of bainite formed isothermally at the same temperature. The bainite will form over a range of temperature, and will display a corresponding range of characteristics.

The bainite reaction may not go to completion, either because of carbon enrichment of the austenite or because the cooling is fast relative to the time required for the completion of the bainite reaction. Austenite remaining may subsequently transform, wholly or in part, to martensite, as the cooling continues.

Martensite

The transformation of austenite to martensite is different from the other reactions by which austenite decomposes in that it does not take place by a process of nucleation and growth, and no diffusion is involved. Rather, lath-like volumes lying along certain crystallographic planes, within the austenite grains, abruptly change, by a shearing mechanism(97-100), from the face-centered cubic austenite structure to the body-centered tetragonal martensite structure. The change in shape of these volumes during the transformation sets up stresses, on a microscopic scale, both in the martensite and in the adjacent untransformed austenite. A photomicrograph illustrating the lath-like structure of the martensite is included in Figure 17. (The "needles" there visible are the intersections of the laths with the plane of polish.)

The martensite reaction starts, on cooling, at a temperature that

appears to be independent of cooling rate(81-86). This temperature is fixed by the thermodynamic equilibrium between face-centered and body-centered structures of the same composition, as modified by the energy involved in setting up the stresses that accompany the transformation(40). In the same way that austenite is unstable with respect to unstrained ferrite of the same composition (bainite) at higher temperatures, it is unstable at low temperatures with respect to strained ferrite (martensite) of the same composition. (The strains are introduced because of the shearing nature of the reaction, and occur both in the martensite and in the surrounding austenite.) Thus a steel, cooled rapidly through the ranges in which the austenite can decompose to ferrite and carbide or to unstressed supersaturated ferrite, can at low temperatures transform to supersaturated ferrite by means of a shearing transformation (not diffusion). ("Upper limit of martensite formation", Figure 9.) The martensite lattice is tetragonal because the carbon atoms do not move during the transition and occupy preferred interstitial positions in the resulting lattice. This tetragonality persists, as the tetragonal lattice is stable at room temperature with respect to supersaturated cubic ferrite having random distribution of carbon atoms, at least when the carbon is above .20 per cent(40).

It is generally believed that once partial transformation has occurred and some stresses are set up, further transformation involves increase in the stresses and hence in the energy associated with them. For sufficient energy to be liberated by the transformation to permit increase of the stresses and the consequent increase of stress energy, the temperature must be lower than that at which transformation started. This stress-energy requirement may be responsible for the outstanding characteristic of the martensite reaction: that it does not take place at constant temperature (except, perhaps, to a very limited extent), but only on cooling.

Much remains to be done in determining the effect of cooling rate upon the progress of the martensite transformation, and even more in determining the factors that govern this effect. There are still some questions as to the M_s temperature itself, for one investigator(87) has reported that very high cooling rates raise the M_s slightly. His results are hardly conclusive, however. The claims(39, 101) that the M_s temperature is raised as the cooling rate is decreased and that the martensite reaction will commence at constant temperature appear to be based upon inferior experimental technique or incorrect interpretation of experimental data, particularly failure to consider tempering. The effects, discussed below, of holding in the proeutectoid, pearlite, and bainite ranges upon the austenite, and hence indirectly upon its subsequent transformation to martensite, also have to be remembered.

Although the M_s temperature, itself, is independent of cooling rate, the amount of martensite formed per degree of cooling below the M_s point is greater the higher the cooling rate(83, 87, 102). The time-temperature diagram for the formation of martensite is shown schematically as Figure 14. An example of the variation with cooling rate of the amount of martensite formed above room temperature is illustrated as Figure 15. The martensite reaction thus differs from the pearlite and bainite reac-

tions, which occur to a lesser degree the higher the cooling rate*. A related effect occurs when austenite is cooled so that it partially transforms to martensite, and is then held at constant temperature for some time. When cooling is resumed the untransformed austenite does not start to transform immediately, but only after a certain amount of cooling has taken place(74, 81, 88, 105, 106, 107). Less of the remaining austenite transforms to martensite on further cooling, to a definite temperature, after the stabilizing treatment then if the cooling is uninterrupted(76, 105-107). The longer the stabilizing treatment(105-107) and the higher its temperature(76) (provided significant tempering does not occur) the greater are these effects.

Several investigators(74, 75, 81, 108) have reported that when cooling is halted after austenite has partially transformed to martensite, the austenite-martensite transformation continues isothermally, for some minutes and to a slight extent, after the cooling has ceased. There is, however, some uncertainty as to the exact time when cooling ceases, and failure in some cases to exclude possible tempering effects. Moreover, the martensite may nucleate transformation of the adjacent austenite to bainite. This local bainite formation, occurring considerably earlier than the general formation of bainite, may be confused with martensite formation. Not much credance, therefore, can be given to the reports of isothermal formation of martensite.

It is not known whether the martensite reaction goes to completion in all cases if the steel is cooled to a sufficiently low temperature. In some steels having a high martensite range, it is generally believed that the reaction does go to completion; for other steels, that have martensite ranges extending below room temperature, it appears that the reaction ceases to progress appreciably, even when cooling is continued(81, 104, 105, 106). This behavior does not seem to have been investigated in detail. It may be that at low temperatures the free energy differences between austenite and martensite does not increase rapidly enough with decreasing temperature to permit the reaction to proceed in the face of the stress-energy opposing it. The reaction ceases or goes to completion gradually or asymptotically, so that it is difficult to determine a definite temperature ("M_f point") below which no reaction occurs, even for a single composition of austenite and a single cooling rate. Since decreasing the cooling rate apparently decreases the temperature for any given per cent transformation, it would be expected that it would decrease the temperature

*It is possible that the effect of cooling rate upon the progress of the martensite reaction arises from the relaxation of relaxation centers(103) set up by the deformation accompanying the reaction. This same relaxation is responsible for the elastic after-effect (Chapter III). Relaxation, although it requires time, occurs rather rapidly. Thus, unless the cooling rate is very high (or the temperatures very low), appreciable relaxation is likely to occur during cooling through the martensite range. Relaxation of the centers increases the stress upon the adjacent material and so increases the amount of cooling necessary for a fixed amount of the remaining austenite to transform to martensite.

for 100% transformation. However, this effect may be obscured by the apparent cessation of the reaction before 100% transformation, and in several types of steel(105-107), even holding at constant temperature within the martensite range has been found to have little effect upon the M_f temperature, though it did decrease the total amount of austenite transformed on cooling to the M_f.

The transformation of austenite to martensite is accompanied by an expansion, which is somewhat greater than those characteristic of the proeutectoid ferrite, the pearlite, and the bainite reactions. Because of the difference between the coefficients of thermal expansion for austenite and for martensite, the amount of expansion increases markedly as the temperature of transformation decreases (Figure 16). This expansion, as will be discussed later, is responsible for the "macro-stresses" that are generally associated with the martensite transformation.

It has often been stated that the initiation and progress of the martensite reaction are appreciably affected by macroscopic stresses, either internal or externally applied. While this may well be the case, no satisfactory measurements seem to have been made. An effect, which has often been stated to exist and has been attributed to stresses (presumably internal macroscopic stresses), is that the M_s occurs at a higher temperature at the surface of a continuously-cooled specimen than in its center, or that on cooling to a temperature within the martensite range more martensite forms at the surface than in the center. In many of the reported cases, however, it does not appear certain that the effect is not due to the difference in carbon content caused by surface decarburization, or to the difference between the mechanical restraints at the boundaries of grains at the surface and of grains within the steel.

The M_s temperature is lower, the higher the carbon content of the austenite, in accordance with thermodynamic considerations(40). For iron-carbon alloys, the temperature falls from about 1000° F. (550° C.) for very low carbon material to about 300° F. (150° C.) for 1.40% carbon and perhaps 150° F. (65° C.) for 1.70% carbon(85, 86). The transition temperature for the transformation of austenite to strained supersaturated ferrite has been calculated(40). The calculated temperatures (Figure 9) fit the experimental data very well. Over a moderate range of carbon content, such as .20 to .90 per cent, the M_s temperature may be considered to vary approximately linearly with carbon content, decreasing about 630° F. (350° C.) for each per cent increase in carbon.

The martensite-start temperature should in general be lowered by those elements that decrease the carbon content of austenite in equilibrium with ferrite. With few exceptions, this seems to be the case. To a first approximation the M_s temperature may be considered to vary linearly with the per cent of an alloying element; the amount it varies, according to the incomplete and somewhat inconsistent data now available, is indicated by the selected values in Table I. Calculations(40) indicate that each element should have a linear effect upon the carbon content associated with a fixed M_s. Since the M_s varies approximately linearly with carbon content over the usual range, little error is introduced by considering

that the elements affect the M_s itself linearly. (The linear relationship should strictly be based upon atomic percentages of the elements, but the use of weight percentages hardly affects its accuracy.) The theory further predicts that the effect of the individual elements upon the lowering of the M_s should be in approximately the same order as their effect in delaying the bainite reaction and lowering its maximum temperature, and in an order not very different from the order of their effect in moving the A_3 line of the equilibrium diagram to the left. These arguments are in reasonable agreement with the experimental data, except perhaps for vanadium and aluminum, on which only one experimental investigation has been made.

The effect of combinations of alloying elements has hardly been studied. One suggestion(113) is that the effects of several elements are additive. It has also been suggested(116) that each element multiplies the M_s temperature (in degrees F.) by a factor that varies linearly with the per cent of the element; because of the use of degrees Fahrenheit, this seems to be a rather arbitrary scheme. The experimental data available do not permit the relative accuracy of the two suggestions to be tested. However, the theoretical work, mentioned above, indicates that the effects of different elements are additive with respect to the carbon content for a fixed M_s temperature. They should, therefore, be approximately additive with respect to the M_s temperature for a fixed carbon content, the earliest and simplest suggestion made.

The preceding discussion of the effects of various elements upon the M_s temperature refers to the amounts of elements in the austenite, and not to the amounts in the steel as a whole. Whether other constituents, such as undissolved carbides or ferrite, affect the M_s temperature of austenite has not been studied, but such an effect seems unlikely. Segregation in the austenite, on a scale comparable to the austenitic grain size or larger, presumably causes the martensite transformation to begin in each grain at a temperature corresponding to the composition of that grain. Segregation in the austenite on a scale small compared to the grain size might raise the M_s slightly or might have no effect.

Very little has been done towards determining the influence of carbon and of alloying elements upon the progress of the martensite reaction. The information available indicates that the M_f temperature is affected in a manner similar to the M_s(111, 113). While there are no experimental data, it would seem that large-scale segregation in the austenite might result in an M_f temperature in each grain corresponding to its composition, while segregation on a scale compared to the grain size might lower the M_f or might have no effect.

There have been reports(86, 117) that raising the austenitizing temperature raises the M_s. No full investigation of this suggestion has been made. If the effect exists, it is very likely due to the effect of austenitizing temperature upon the grain size, since it is possible that increasing the grain size raises the M_s (although the contrary has been suggested(84)).

At a given temperature, changes in carbon content and in alloy content affect the specific volume of austenite to about the same extent that they affect the specific volume of martensite(118). Thus, for constant temperature of transformation, the volume change from austenite to martensite is practically independent of composition. If, however, a difference in composition results in a difference in temperature of transformation, there will be a corresponding difference in volume change (Figure 16).

Relations Among Stages

The relations among the proeutectoid, pearlite, and bainite reactions discussed in the section on isothermal transformation apply also to continuous cooling. The applications of the relations are more important in the case of continuous cooling than of isothermal transformation, for when austenite is continuously cooled it ordinarily does not reach the bainite temperature range until it has been in the pearlite range for a significant time. As previously mentioned, the effect of holding austenite in the pearlite range upon its subsequent transformation to bainite has as yet been investigated only sketchily.

It remains to consider the relations between the martensite transformation and the proeutectoid, pearlite, and bainite transformations. The effect of partial transformation to martensite upon subsequent transformation of the remaining austenite to proeutectoid constituents or to pearlite has been studied only scantily(76, 111). Since the proeutectoid and pearlite reactions ordinarily do not occur before the bainite reaction at the temperature at which martensite forms, the effect of martensite upon them must be investigated by reheating the metal after some martensite has formed. The primary effect of the martensite is probably to nucleate and consequently hasten(88) the high temperature reactions.

The weight of evidence indicates that partial transformation to martensite considerably hastens transformation of the remaining austenite by the bainite reaction(78, 89). This occurs both when a steel is cooled to a temperature within the martensite range and held there (which has confused several investigators of the decomposition of austenite), and when the steel, after being cooled within the martensite range, is reheated to a higher temperature within the bainite range. Presumably the phenomenon also occurs during continuous cooling.

Studies of the effect of partial transformation in the pearlite range upon the M_s temperature of the remaining austenite have been made(74, 75, 76) upon several high-alloy steels, mostly of high carbon contents. They found that the M_s temperature was raised when partial transformation occurred, and also when the austenite was held in the pearlite range for a time insufficient for any detectable transformation. They attribute this effect to precipitation of carbides and consequent lowering of the carbon content of the remaining austenite. If this is correct, an effect of proeutectoid carbide precipitation, rather than of pearlite formation, was presumably being investigated. It then appears that no investigations of the effect of partial transformation to pearlite have been made; there may

well be no effect. Neither have any studies of the effect of partial transformation to proeutectoid ferrite been made; presumably the remaining austenite would be enriched in carbon and its M_s temperature consequently lowered.

As indicated earlier, during the transformation to bainite the remaining austenite becomes enriched in carbon. The M_s temperature of the remaining austenite is lowered because of its higher carbon content.

Retained Austenite

The term "retained austenite" refers simply to the austenite remaining untransformed in a steel after completion of the cooling and isothermal holding. Since whatever austenite does not transform by the proeutectoid, pearlite, bainite, and martensite reactions is "retained", little separate consideration of the factors influencing its presence is necessary.

Retained austenite is commonly of interest in conjunction with an otherwise martensitic structure. In this case the controlling factor is the relation between the martensite temperature range and the minimum temperature to which the steel is cooled, although other features of the martensite transformation (such as the degree of completion of the reaction upon cooling below the M_f and the effect of cooling rate) also play a part. In the majority of cases, steel is not cooled below room temperature after quenching; if the M_f lies below room temperature, some austenite will then necessarily be retained. If the steel is not cooled as low as room temperature, the likelihood of retained austenite and the amount retained are greater.

Tempering

Tempering of the Decomposition Products of Austenite

The various decomposition products of austenite themselves change progressively at temperatures below the A_1, tending towards the equilibrium condition* of a single globule of carbide in a matrix of ferrite. These changes are called "tempering". Within part of the temperature range in which tempering occurs, another constituent also precipitates in many steels. The precipitate, not yet positively identified, is responsible for the phenomenon of temper brittleness.

Tempering of Martensite

Let us consider first the tempering of martensite in iron-carbon alloys. Iron carbide tends to precipitate, reducing the carbon content of the martensite. There is no reason to doubt that this precipitation occurs by a process of nucleation and growth, and closely resembles, in its mechanism and kinetics, the precipitations from solid solution that are respon-

*Excluding graphitization.

sible for "aging" in non-ferrous alloys. Little basic work has been done on the precipitation of carbide from martensite, but nucleation evidently takes place at many points within each martensite crystal. Once formed, the nuclei grow into carbide particles of unknown shape. (The particles are too small to be observed with the light microscope, and investigation by more suitable techniques has not yet been made.) The lattice structure of the particles also has not been determined. It is, therefore, not established whether cementite precipitates directly or, as in many other systems, a transition structure precipitates first(106) that later transforms into cementite. The precipitation is accompanied by a small decrease in volume.

Most of the carbon precipitates in a matter of seconds at temperatures of 400° F. (200° C.) or higher, of minutes at 200° F. (100° C.), and of perhaps years at room temperature(119). The variation of the rates of nucleation and of growth with temperature has not been considered either theoretically or experimentally. The rates must depend on the free energies of the carbide and of the martensite (as a function of its carbon content), the interfacial tension between carbide and martensite, and the rate of diffusion of carbon in martensite.

The martensite, unless it is of low carbon content, will be tetragonal before tempering. The tetragonality, as has been mentioned, is caused by the presence of the carbon atoms in a group of preferred positions in the martensite lattice. At carbon contents above about .20 per cent at room temperature and .35 per cent at 400° F. (200° C.), the tetragonal structure is thermodynamically more stable than a cubic structure having the carbon atoms distributed randomly(40). As carbon precipitates from martensite during tempering, the carbon content of the martensite is reduced below the amounts mentioned, and the cubic structure becomes stable with respect to the tetragonal. Transformation of the tetragonal to the cubic structure requires only the movement of carbon atoms from one lattice position to an adjacent one. This transformation, therefore, occurs very rapidly as compared with the carbide precipitation, which involves carbon diffusion over considerably greater distances. It may be noted that, according to theory(40), the transformation of tetragonal to cubic martensite should be reversible. For example, martensite tempered at 400° F. (200° C.) down to a carbon content of .30 per cent, and therefore having a cubic structure, should on cooling to room temperature revert to the tetragonal structure.

Transformation of tetragonal martensite to cubic does not interfere with the precipitation of iron carbide. On continued tempering, carbide precipitates from the cubic martensite and the carbon content of the latter is reduced towards the equilibrium value for solubility of carbon in ferrite. Thus, the cubic martensite becomes simply ferrite.

When the carbon content of the martensite has been sufficiently depleted, spheroidization of the carbide particles and growth of the larger particles(30, 121) at the expense of the smaller commence. The interface energy is thus reduced by decreasing the area of interface between carbide and ferrite (martensite). During the process, the carbide dissolves in the ferrite where the radius of curvature of a carbide particle is small. The

resultant dissolved carbon diffuses through the ferrite, and precipitates out as carbide where the radius of curvature of a carbide particle is large. Undoubtedly transformation to cementite of the transition iron carbide, if any, takes place before this process proceeds very far. A small decrease in volume, presumably accompanying the transformation, has been observed. This volume change forms the chief evidence for the existance of a transition carbide(119).

Precipitation, transformation to cementite of the transition carbide (if any), and spheroidization and growth of the carbides probably all overlap. Appreciable spheroidization and growth probably occur within a few hours at 200° F. (100° C.) and a few seconds at 550° F. (300° C.). Empirically(122), it has been found that the amount of spheroidization and growth (as measured by hardness) depends on the parameter, $T \log (t/t_0)$, where T is the absolute temperature, t the time, and t_0 a constant that apparently depends upon the carbon content of the steel, varying from the order of 10^{-16} seconds for 0.30 per cent carbon to 10^{-12} seconds for 1.20 per cent carbon. Photomicrographs illustrating the precipitation and growth of the carbide during tempering of martensite are included in Figure 17.

The original martensite crystals probably remain as individual crystals (of ferrite) throughout the tempering process. Since the grain size of the martensite probably increases with increasing grain size of the austenite from which it formed, the grain size of the ferrite in the tempered structure probably increases with increasing grain size of the original austenite. At high tempering temperatures and long tempering times, ferrite grain growth may occur. If the ferrite (or martensite) has been deformed plastically, recrystallization may also take place.

Tempering of Bainite

If bainite, as formed, contains ferrite supersaturated with carbon, this carbon will precipitate upon tempering. If during the formation of the bainite the rate of carbon diffusion was sufficiently rapid relative to the rate of sidewise growth of supersaturated ferrite, carbide plates, or at least platelike regions of high carbon concentration, will have formed. During tempering, carbon from supersaturated ferrite will then tend to precipitate along these plates or platelike regions. If the bainite is formed under such conditions that the structure consists wholly of homogeneous supersaturated ferrite, the tempering will be essentially the same as that of martensite (except for differences arising from the lower internal stresses).

When the carbon content of the ferrite has been lowered sufficiently, spheroidization and growth of the larger cementite particles at the expense of the smaller commence, just as in the tempering of martensite. If the original iron carbide precipitate had the structure of some transition phase rather than of cementite, tempering will soon cause transformation of the transition structure to cementite. The non-uniformity of the original precipitate persists, however, for some time; the particles tend to line up in planes, presumably the sites of the original carbide plates or high-carbon regions. As the particles become larger and fewer,

tempered bainite approaches the structure of tempered martensite. To distinguish the two under the microscope is difficult. The empirical relation between temperature and time found for spheroidization and growth in the tempering of martensite seems to apply to the tempering of bainite also.

The original crystals of supersaturated ferrite presumably remain as individual crystals throughout the tempering process. If the size of these original crystals increases with increasing grain size of the austenite from which they formed, the grain size of the ferrite in the tempered structure also increases with the grain size of the original austenite. At high tempering temperatures and long tempering times growth and recrystallization of the ferrite grains may occur, just as in the case of martensite.

Tempering of Pearlite

Since pearlite consists of ferrite and carbide of practically equilibrium carbon concentrations, there are no stages in its tempering involving loss of tetragonality or precipitation of appreciable amounts of cementite. Since, as indicated in Figure 9, the ferrite may be slightly supersaturated with respect to carbon, tempering can cause a slight precipitation of cementite or other carbide. Primarily, however, tempering of pearlite is a process of spheroidization and of growth of large carbide particles at the expense of small. Each carbide lamella tends to break up into a group of spheroids, so that the distribution of carbides in tempered pearlite is much less uniform than in tempered martensite and somewhat less than in tempered bainite. As the time and temperature increases, and the particles become larger and fewer, the structure approaches those of tempered martensite and tempered bainite. There is some evidence that the relation between temperature and time in the tempering of pearlite is the same as that in the spheroidization of tempered martensite.

The ferrite in certain regions of each nodule (pearlite patch grown from a single nucleus) is of the same orientation(45), and on tempering it remains as a single ferrite grain. Since the size of the pearlite nodules increases with increasing size of the austenite grains, the ferrite grain size of the tempered pearlite probably increases with increase of the original austenitic grain size. Grain growth and recrystallization may occur at high tempering temperatures and long tempering times.

Tempering of Proeutectoid Constituents

Proeutectoid ferrite in iron-carbon alloys is little affected by tempering. If the carbon concentration in the ferrite is higher than the equilibrium concentration at the tempering temperature, precipitation of cementite will occur. This tends to take place around the edges of the proeutectoid constituent, particularly at cementite particles formed there from other constituents. Carbide particles formed around the edges will then grow into the originally proeutectoid region. If the carbon concentration of the ferrite is lower than the equilibrium concentration at the tempering temperature, diffusion of carbon from neighboring constituents into the proeutectoid ferrite will occur, provided the temperatures and times are high enough to permit carbon diffusion over the relatively long

distances involved.

Proeutectoid cementite in iron-carbon alloys can only spheroidize during tempering. At sufficiently high temperatures and times grain-boundary layers of cementite will break up into groups of spheroids. These particles will be similar (except perhaps in size) to cementite spheroids formed from other constituents. The largest spheroids from either source will grow and the smaller redissolve.

Effect of Carbon Upon Tempering

Increasing the amount of carbon in the steel increases the amount of the carbides in the tempered structures. It also increases the total amount of carbon that must diffuse in order for the various changes to take place. Increase in carbon, therefore, probably delays the tempering somewhat.

Effect of Alloying Elements Upon Tempering

Alloying elements would be expected to have only small effects upon the processes or rates of precipitation of iron carbide from martensite, of transformation from tetragonal martensite to cubic, of transformation (if any) from transition iron carbide to cementite, and of spheroidization and growth of cementite. However, practically no experimental data are available for any of these reactions.

When alloying elements are present, an additional stage of tempering, not found in iron-carbon alloys, is likely to take place. This transformation consists of change in the alloy contents of the ferrite and carbide phases, and involves diffusion of the alloying element through ferrite, and to some extent through carbide. For this reason, it takes place to an appreciable extent only at moderately high tempering temperatures and tempering times.

When carbide-forming elements are present, change in alloy content of the carbide may involve not merely change in cementite composition, but also change in the crystal structure of the carbides(76, 123). This transformation apparently takes place by a process of nucleation and growth that in general involves diffusion of both carbon and alloying elements toward the newly-forming carbides. The cementite particles are not likely themselves to serve as nuclei for complex carbides, but these carbides will tend to nucleate at the interface between cementite and ferrite. As the complex carbides grow, they will tend to absorb neighboring cementite. Solution of cementite in the ferrite and simultaneous precipitation of carbon from ferrite upon complex carbides can also take place; this process is a kind of spheroidization. Whether or not the cementite will eventually disappear entirely depends upon the equilibrium field within which the steel lies at the temperature of tempering. Spheroidization of the complex carbides themselves and growth of the larger particles at the expense of the smaller will take place when the ferrite is sufficiently depleted in the alloying elements.

Alloying elements affect the tempering of bainite in the same way

that they affect the tempering of martensite. Their effect upon the tempering of pearlite is different only insofar as the carbides forming the pearlite lamellae may be of a structure other than cementite. There is reason to doubt that complex alloy carbides are present in untempered pearlite. However, if they are, the particles of complex carbide existing before tempering will serve as nuclei for further formation of the complex carbide at the expense of whatever cementite may be present.

Proeutectoid carbides in an alloy steel may certainly consist partly or wholly of complex carbides rather than cementite. Particles of complex carbide will, on tempering, grow at the expense of the cementite, if any is present. Simultaneously, but more slowly, the particles of complex carbide will spheroidize and the larger particles will grow at the expense of the smaller.

There is some evidence(122) that the empirical relationship between the time and the temperature that produce a given degree of tempering in the case of precipitation and growth of cementite applies also to precipitation and growth of complex carbides.

General

The term tempering has been used above in the broad sense of changes with time of the decomposition products of austenite (other than austenitizing and graphitizing). This differs somewhat from the narrower sense commonly used in heat-treatment, reheating of quenched steel to a temperature below the A_1. The changes in the decomposition products of austenite resulting from cooling to room temperature and then reheating are not appreciably different from those resulting when the products are brought directly from the temperature at which they formed to the tempering temperature. The decomposition products of austenite begin to change (and so to temper in the broader sense) as soon as they are formed. For example, if pearlite is forming isothermally, the first pearlite formed will have undergone some tempering while other pearlite is yet to form from the austenite. And, likewise, if austenite is transforming to martensite during quenching (rapid continuous cooling), the martensite formed at high temperatures will temper to some extent before all the martensite has formed.

Decomposition of Retained Austenite on Tempering

If a quenched steel is reheated to a temperature below the A_1 range, any austenite remaining after the quench will tend to decompose during the reheating and subsequent holding at temperature. The decomposition will take place as if the austenite were brought directly to the temperature in question(111), except that the decomposition products present before reheating tend to nucleate subsequent decomposition. Thus, further decomposition will begin more quickly than if no previous decomposition had taken place.

Temper Brittleness*

In addition to the series of changes in the decomposition products of austenite described above (which include transformation of tetragonal martensite to cubic, precipitation, growth, and spheroidization of iron carbide, and precipitation, growth, and spheroidization of alloy carbides), another type of precipitation from ferrite may occur at temperatures below the critical. This precipitation does not seem to take place in iron-carbon alloys or plain-carbon steels; it has been reported only in steels containing more than .60% manganese or appreciable percentages of chromium or nickel. The precipitation is apparently associated with the decreasing solubility of some constituent in ferrite with decreasing temperature. (Figure 18.) The precipitation is similar to the other well-known cases of precipitation from solid solution, such as those encountered in aging. At any temperature below that at which the amount of constituent present is soluble in the ferrite, precipitation will occur, given sufficient time, and will continue until the concentration of the constituent in the ferrite is reduced to the equilibrium value for that temperature. Since this equilibrium value (the solubility) decreases with decreasing temperature, the total amount of precipitation that can take place increases with decreasing temperature.

Apparently no precipitation occurs above about 1100° F. (600° C.), regardless of the initial composition of the ferrite. If, as indicated in Figure 18, this maximum temperature were the temperature at which the solubility of the constituent in ferrite is equal to the initial concentration of the constituent in ferrite, it would be expected to vary appreciably with the composition of the steel, for the initial concentration of the constituent would be expected to vary. Since this maximum temperature does not vary appreciably from steel to steel, it seems likely that the temperature 1100° F. (600° C.) is represented by some horizontal line in the equilibrium diagram, such as the ternary eutectoid line indicated in Figure 19.

The precipitation proceeds by a process of nucleation and growth. Nucleation appears to take place preferentially at the prior austenitic grain boundaries, where the interface energy that must be supplied to form a nucleus is least**. The particles of precipitate remain small; they have not yet been found with certainty by microscopic examination. The total amount of precipitate is also small, but, as mentioned above, increases as the temperature is decreased.

The precipitation takes place slowly at temperatures just below 1100° F. (600° C.), more rapidly at lower temperatures and slowly again at

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*The fundamental work on this subject was done by Greaves and collaborators(124-128). A review(129) has recently been published summarizing the available information.

**Why less energy should be required at prior austenitic grain boundaries than at ferrite grain boundaries is not clear; perhaps very small inclusions are left at the austenitic grain boundaries after the austenite has decomposed.

still lower. (See Figure 18.) This behavior, determined by the variation with temperature of the solubility, the diffusion rate of the responsible constituent, and the surface energy of the precipitate, is usual in reactions taking place by nucleation and growth. The precipitation ceases soonest at high temperatures, where the amount of constituent that can form is smallest.

If precipitation has occurred, and the steel is held at a temperature at which the solubility of the constituent in ferrite is greater than the amount in solution, the precipitate will start to dissolve in the ferrite. Solution will continue until the concentration of the dissolved constituent reaches the solubility limit or the precipitate is entirely dissolved. The solution appears to take place more rapidly the higher the temperature. This might be expected, as no nucleation is necessary for the process (the ferrite already being present).

The constituent responsible for temper brittleness appears to be soluble in austenite in the concentrations in which it occurs in steel, and there is no evidence that the precipitation can take place from austenite.

The natures of the temper brittleness precipitate and of the constituent whose solubility in ferrite governs the precipitation have not yet been definitely established. There are indications that the precipitate is iron nitride, its precipitation being governed by the solubility of nitrogen in ferrite (in the presence of carbon and of whatever alloying elements may be present).

As mentioned above, the temper brittleness precipitation does not seem to take place in plain-carbon steels. There is reason to believe that as the amounts of manganese, nickel, and chromium are increased, the amount of precipitation that can take place also increases. The same may be true for molybdenum, but this element markedly decreases the rate of precipitation.

The temper brittleness precipitation can take place during all forms of tempering in the broad sense, below the equilibrium temperature for precipitation. Thus, precipitation occurs during holding at the tempering temperature and during cooling from this temperature. It can also take place in ferrite newly formed from austenite, while this ferrite is still at the temperature at which it formed, as well as while the ferrite is cooling from that temperature.

TABLE I

EFFECT OF CARBON AND ALLOYING ELEMENTS ON M_s TEMPERATURE

Element	Effect Per % of Element °C.*	°F.**
C (.20 - .90%)	-350	-630
Mn	- 40	- 72
V	- 35	- 63
Cr	- 20	- 36
Ni	- 17	- 31
Cu	- 10	- 18
Mo	- 10	- 18
W	- 5	- 9
Si	0	0
Co	+ 15	+ 27
Al	+ 30	+ 54

Note: The M_s temperature for iron-carbon alloys, extrapolated to 0 per cent carbon, is approximately 550° C. (1010° F.)*.

*Selected from the literature(86, 111-115).
**Converted from ° C.

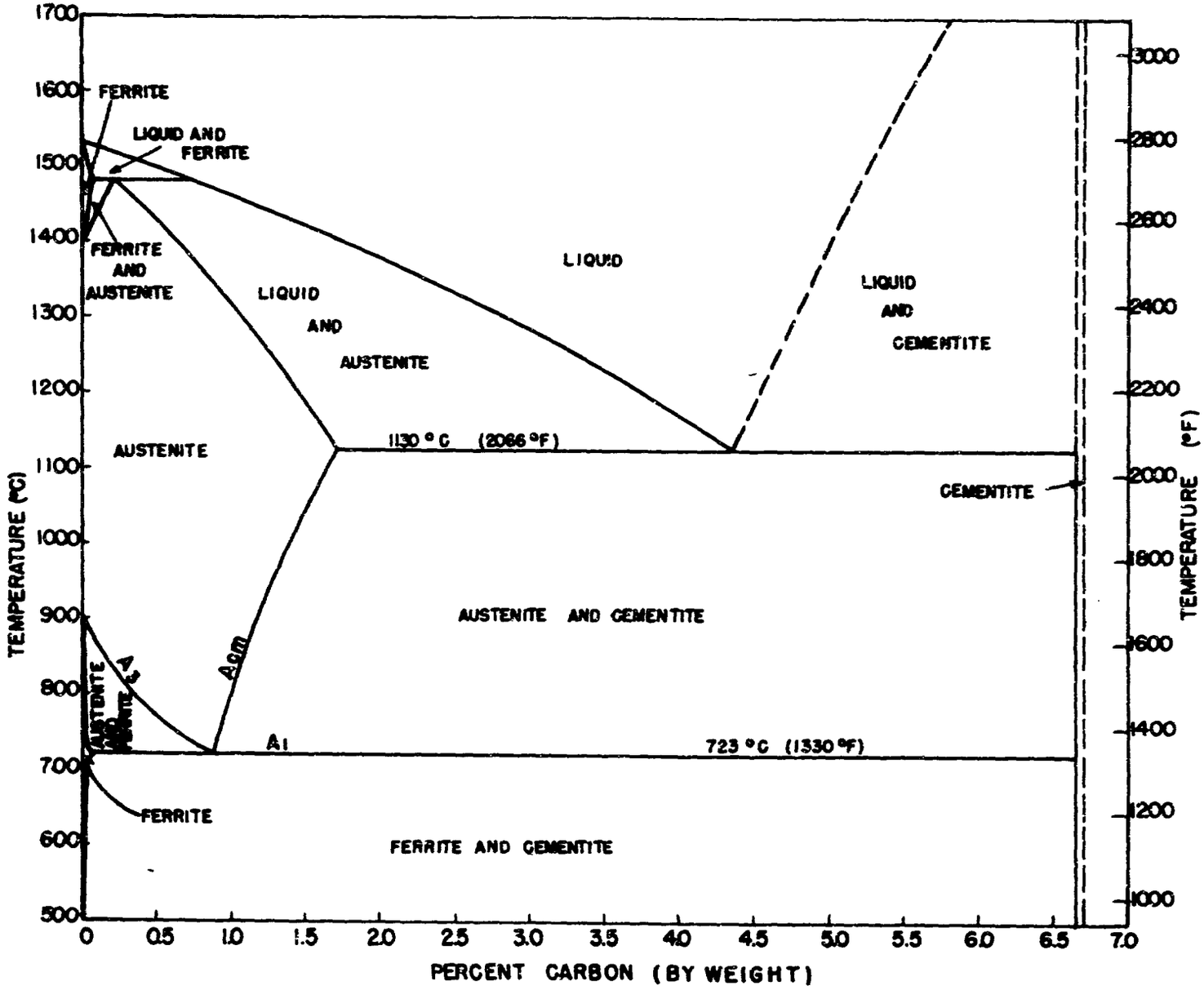


FIGURE I

IRON-CARBON EQUILIBRIUM DIAGRAM. (FROM DATA SELECTED BY EPSTEIN[1])

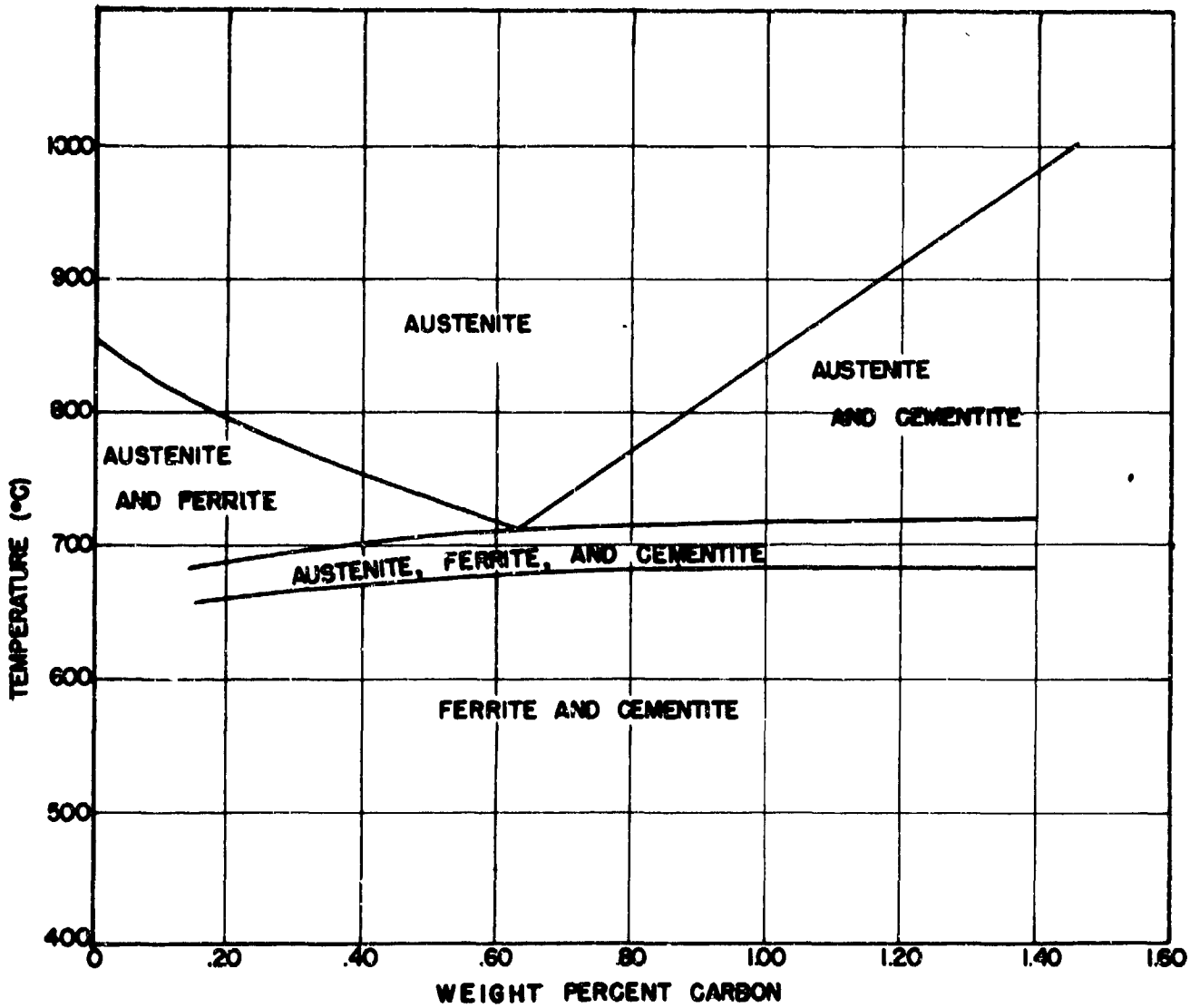


FIGURE 2

2.5% MANGANESE SECTION OF THE TERNARY EQUILIBRIUM DIAGRAM FOR ALLOYS OF IRON, MANGANESE AND CARBON. (AFTER WELLS[4])

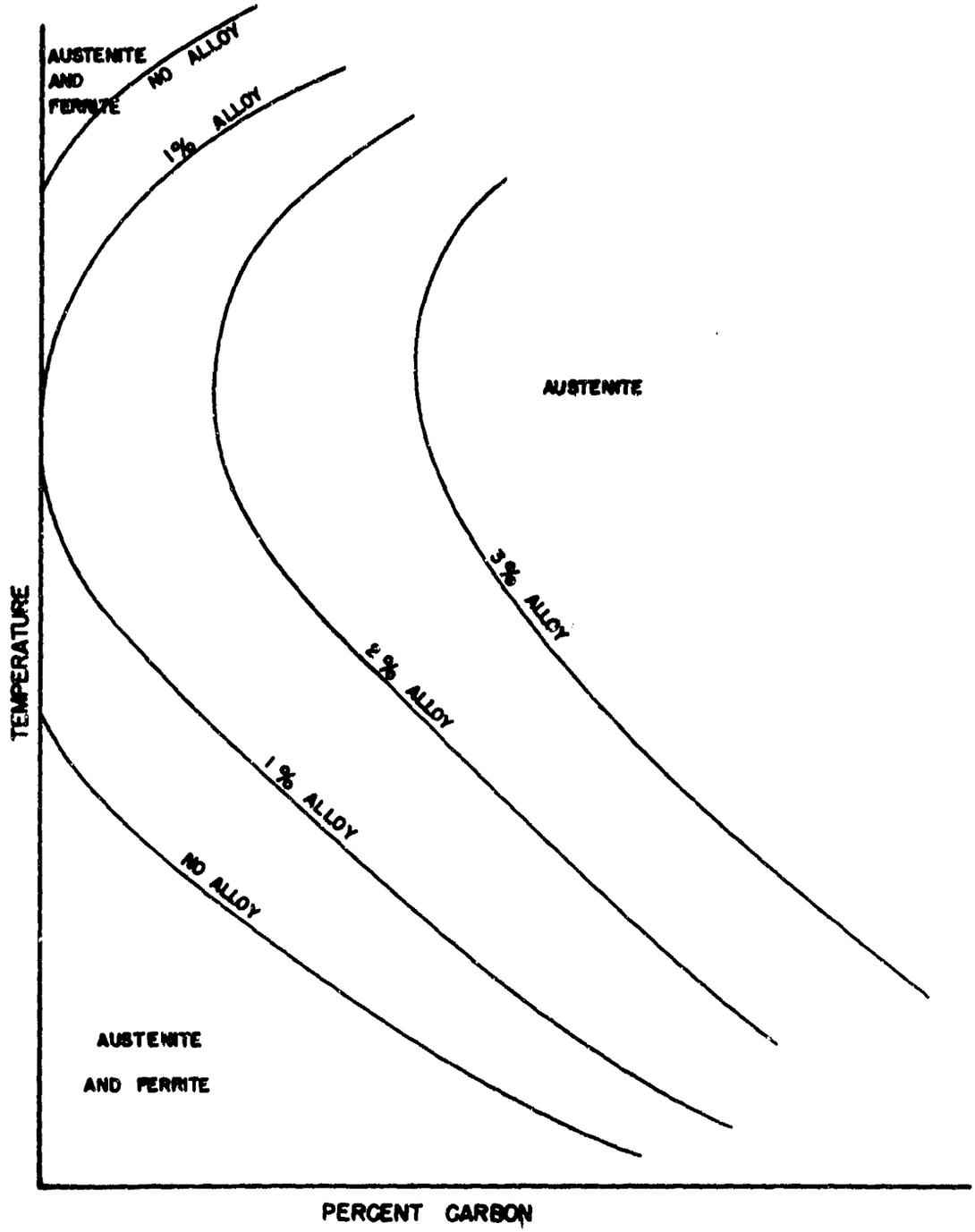
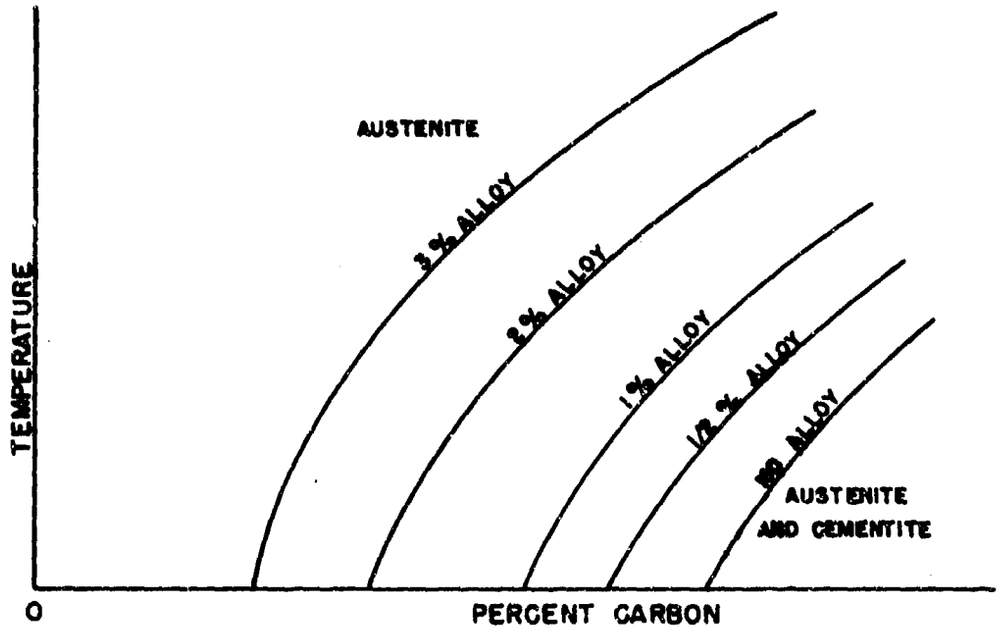


FIGURE 3

SHIFT OF A_3 LINE TO HIGHER CARBON CONTENTS BY ADDITIONS OF AN ALLOYING ELEMENT (SCHEMATIC)



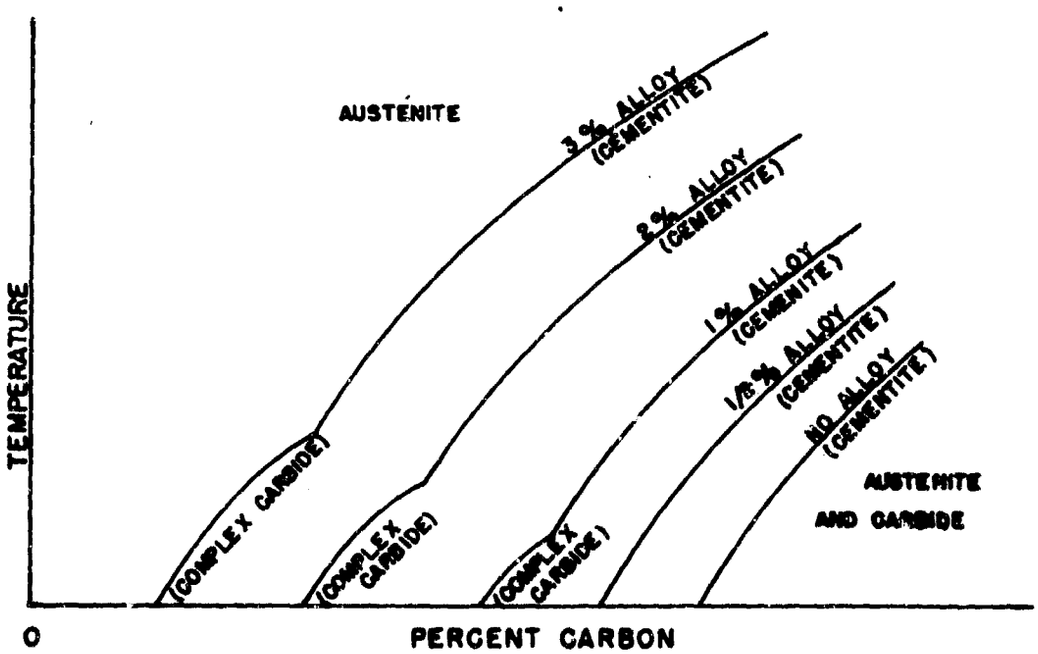


FIGURE 4

EFFECT OF ALLOYING ELEMENT UPON Acm LINE
(SCHEMATIC) (AFTER CONCEPT OF ZENER[2])

TOP - NO COMPLEX CARBIDE FORMED
BOTTOM - COMPLEX CARBIDE FORMED

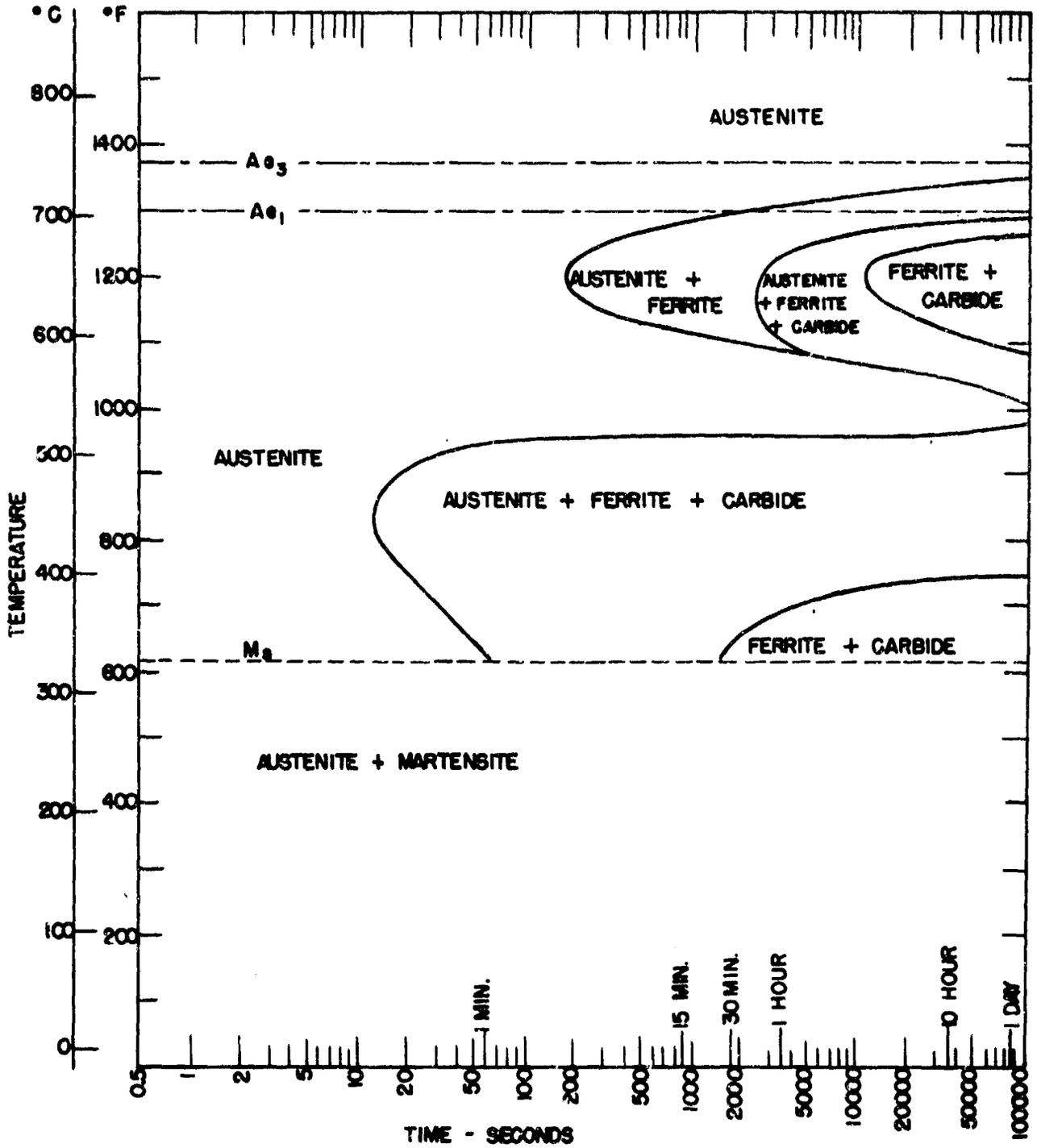


FIGURE 5

TYPICAL ISOTHERMAL TRANSFORMATION DIAGRAM FOR STEEL CONTAINING CARBIDE FORMING ELEMENTS (SAE 4340)

(AFTER ATLAS OF TRANSFORMATION DIAGRAMS[38], MODIFIED)

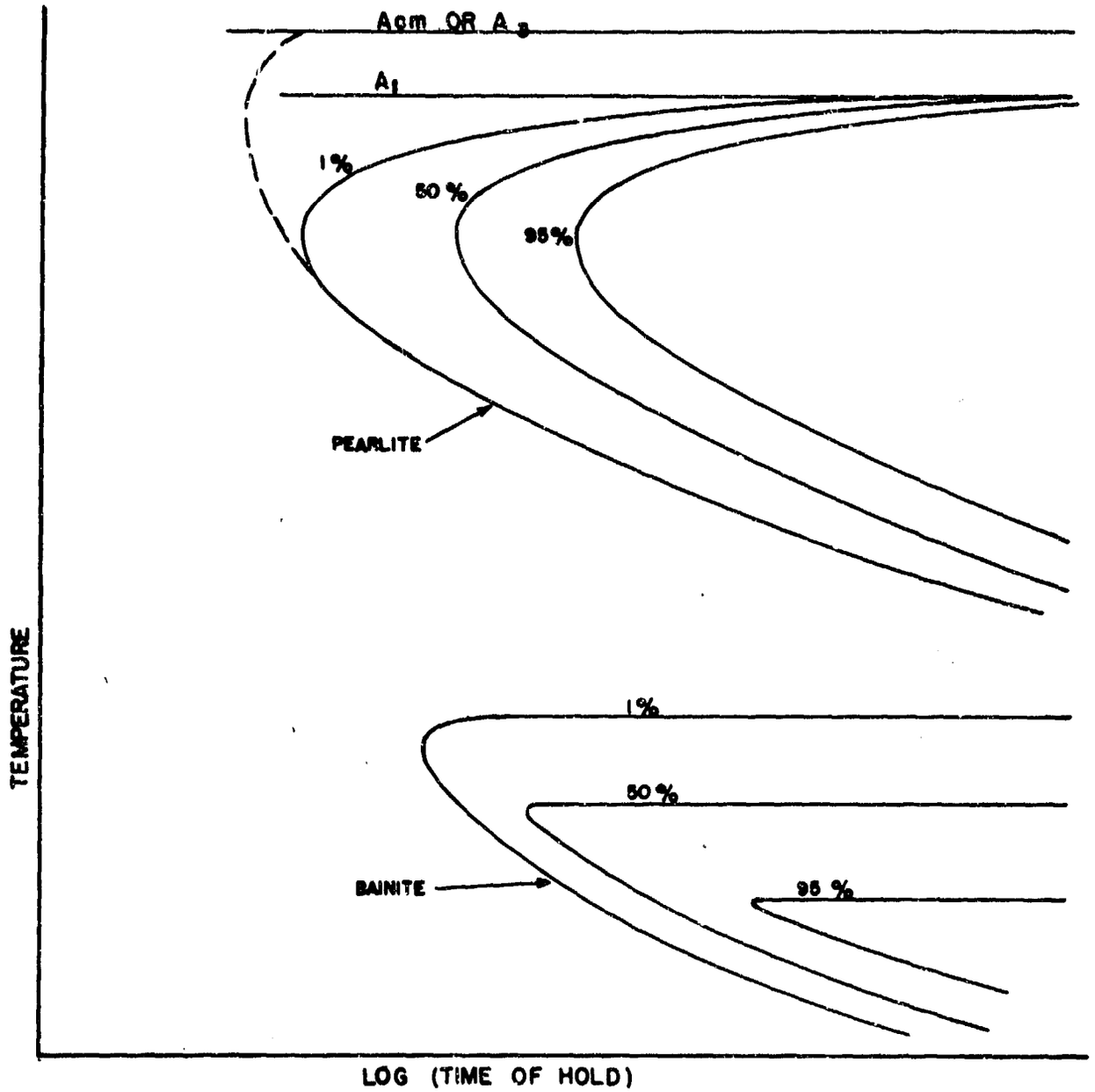


FIGURE 6

IDEALIZED TRANSFORMATION DIAGRAM FOR ISOTHERMAL DECOMPOSITION OF AUSTENITE (AFTER ZENER,[40] SLIGHTLY MODIFIED)

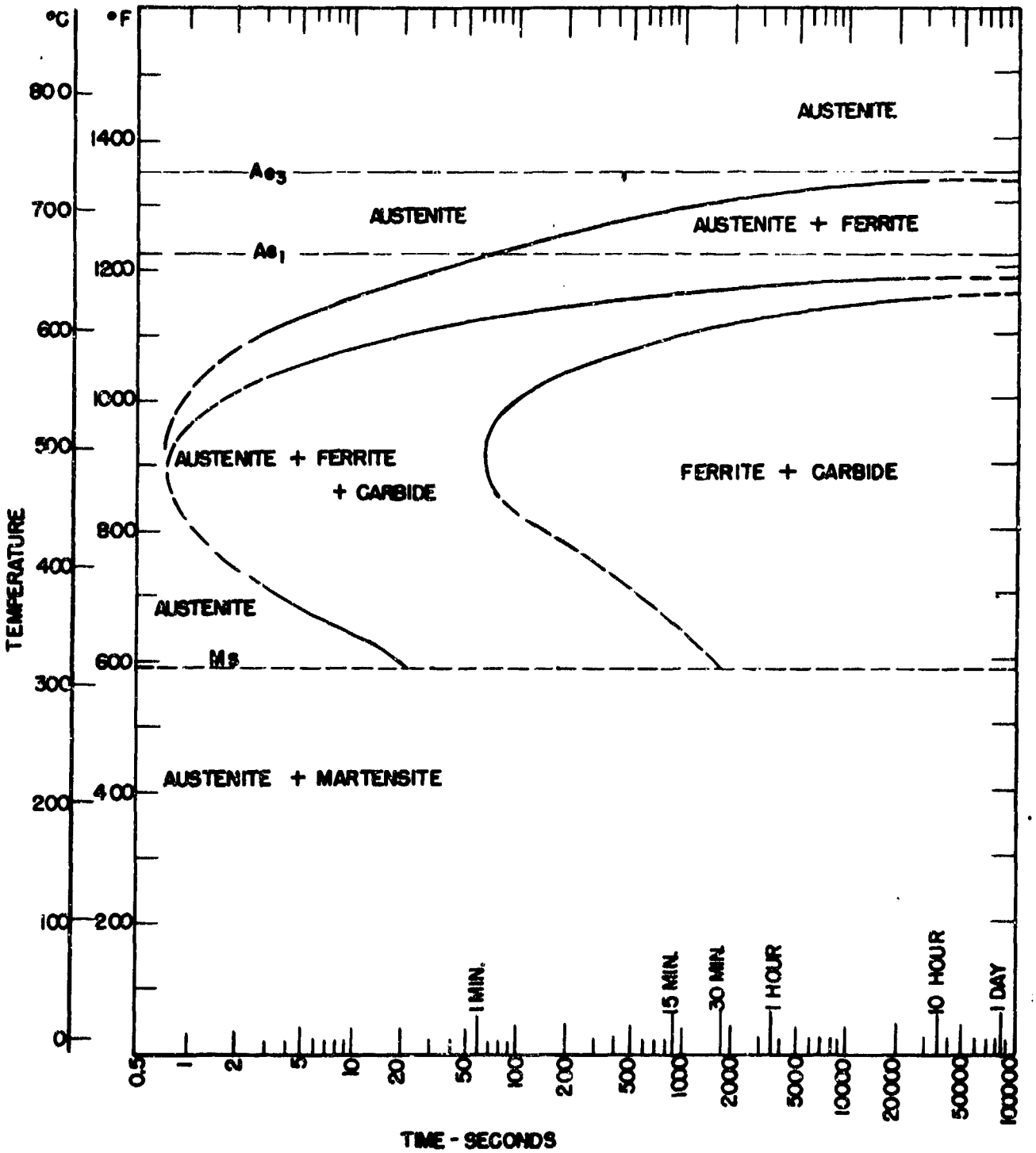


FIGURE 7

TYPICAL ISOTHERMAL TRANSFORMATION DIAGRAM FOR STEEL NOT CONTAINING CARBIDE-FORMING ELEMENTS (SAE 2340)

(AFTER ATLAS OF TRANSFORMATION DIAGRAMS[39], MODIFIED)

FIGURE 8

PHOTOMICROGRAPHS OF PEARLITE

AND PROEUTECTOID PRODUCTS

A. Lamellar Pearlite (X 2000)

B. Nodules of Fine Pearlite at Prior Austenitic Grain Boundaries (X 1000)

C. Proeutectoid Ferrite at Prior Austenitic Boundaries Surrounding Fine Pearlite (X 1000)

D. Proeutectoid Cementite at Prior Austenitic Grain Boundaries Surrounding Coarse Pearlite (X 2000)

B

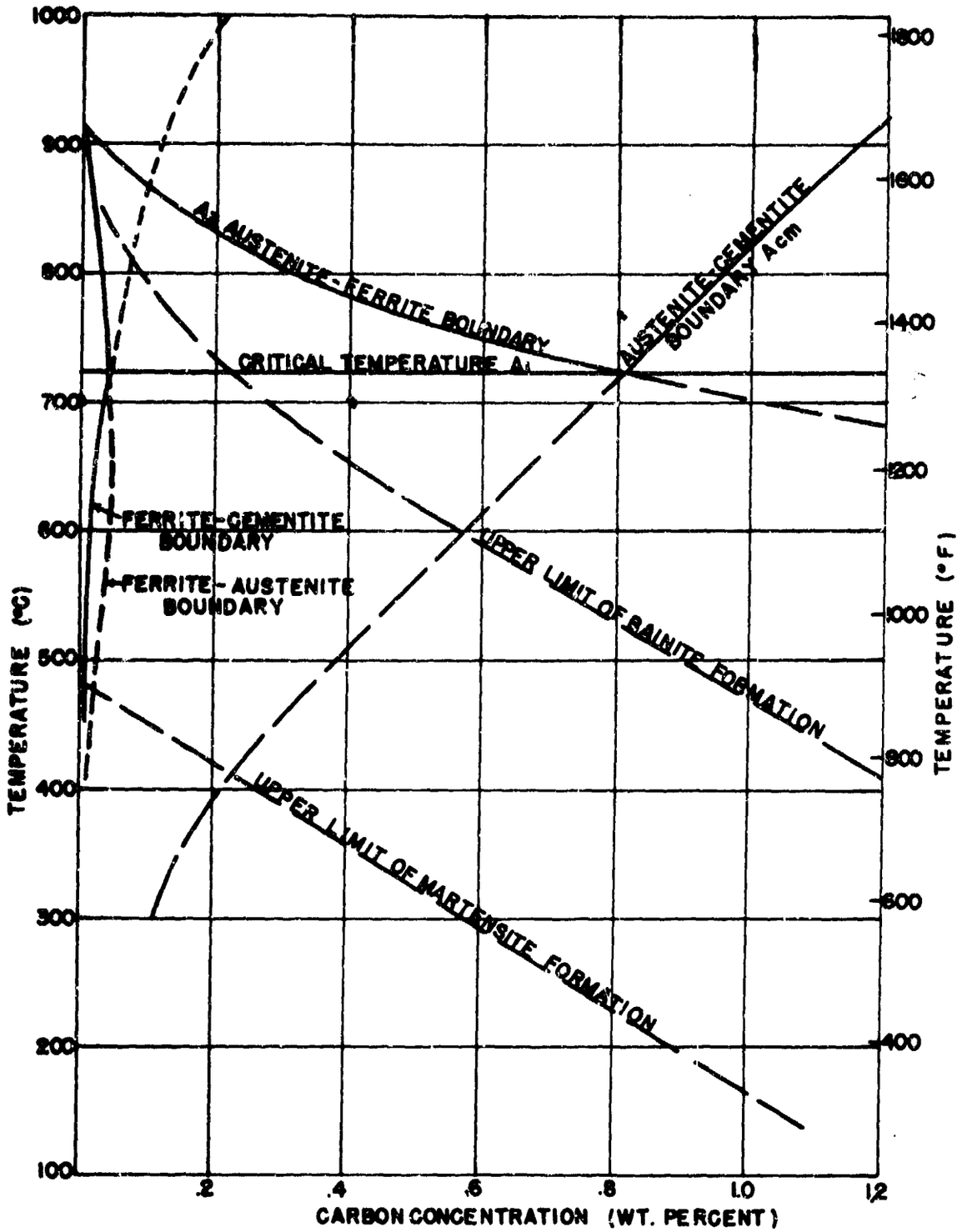


FIGURE 9

QUASI-EQUILIBRIUM RELATIONS IN THE IRON-CARBON SYSTEM (FROM ZENER[40])

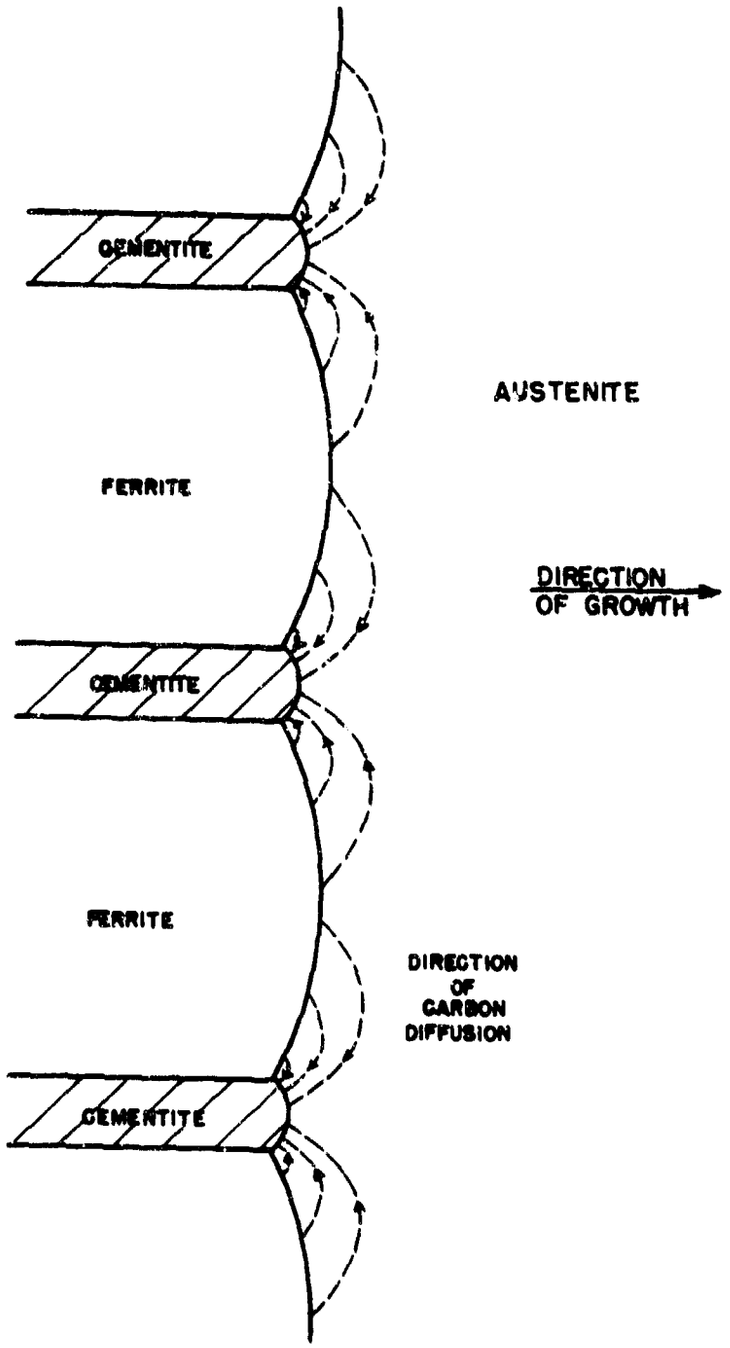


FIGURE 10

DIRECTION OF CARBON DIFFUSION DURING EDGEWISE GROWTH OF PEARLITE (SCHEMATIC)

FIGURE 11

PHOTOMICROGRAPHS OF BAINITE

A. A Needle of Bainite in a Matrix of Martensite (X 2500)

B. Several Needles of Bainite in a Matrix of Martensite (X 2500)

C. Bainite Having Carbide Precipitate at Needle Boundaries (X 2500)

D. Bainite Having Random Carbide Precipitate (X 2500)

A

B

D

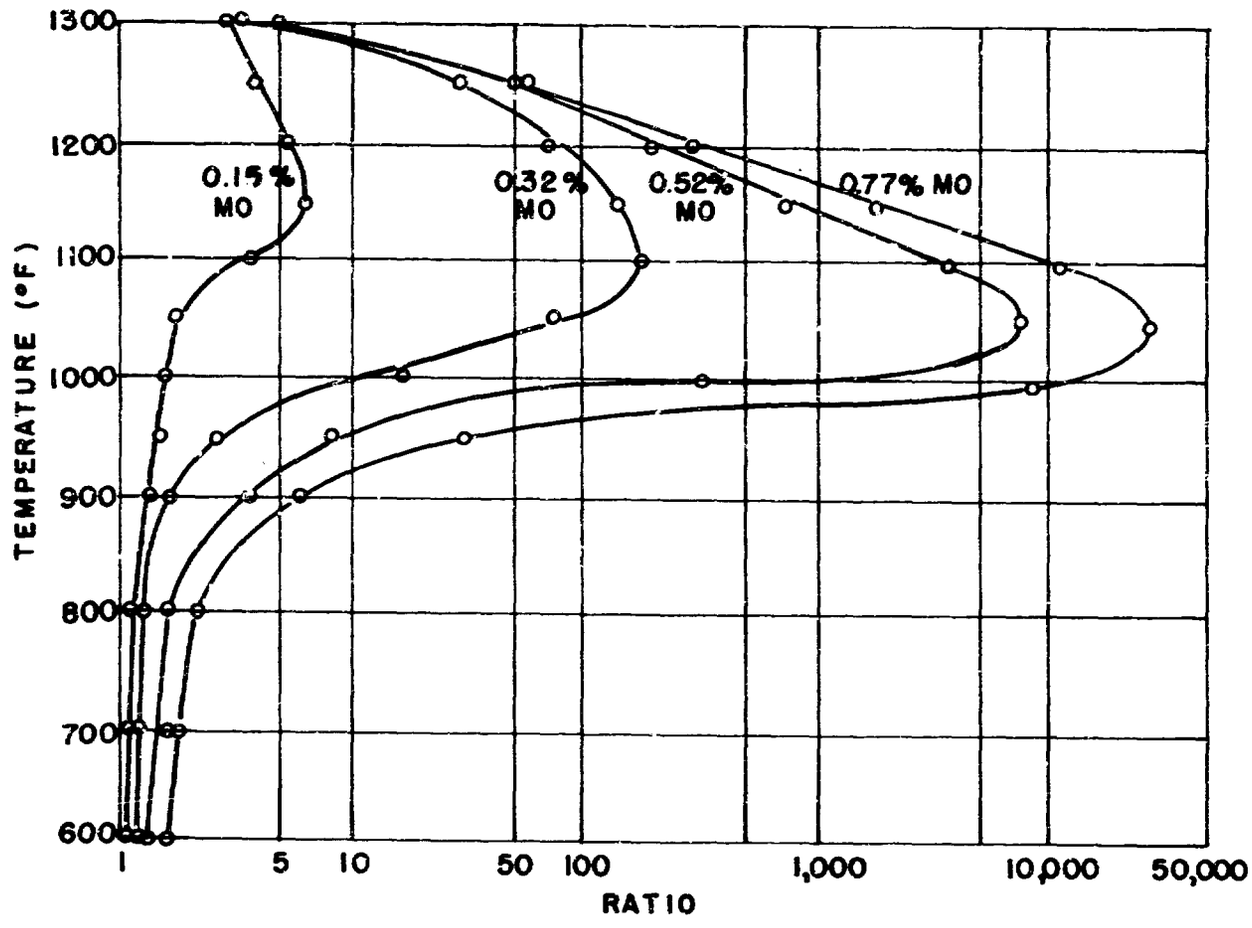


FIGURE 12

EFFECT OF MOLYBDENUM UPON ISOTHERMAL DECOMPOSITION OF AUSTENITE.

RATIO OF TIME FOR TRANSFORMATION OF MOLYBDENUM STEELS TO THE TIME FOR TRANSFORMATION OF PLAIN CARBON STEELS. EUTECTOID STEELS.

(AFTER BLANCHARD, PARKE & HERZIG[69])

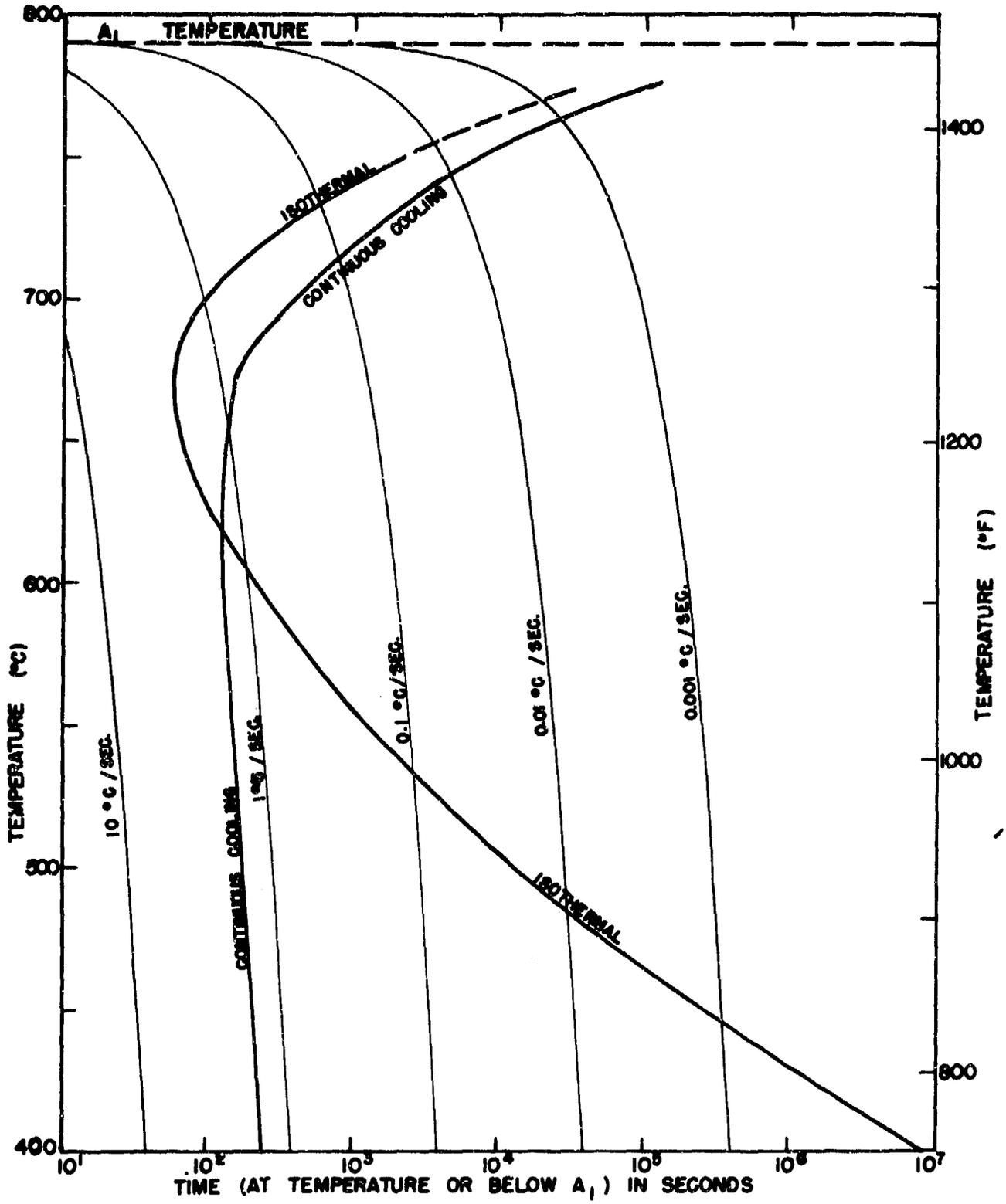


FIGURE 13

TYPICAL ISOTHERMAL DIAGRAM AND DERIVED CONTINUOUS COOLING DIAGRAM FOR INITIATION OF TRANSFORMATION FROM AUSTENITE TO PEARLITE. (CONTINUOUS COOLING DIAGRAM DRAWN FOR CONSTANT COOLING RATES. COOLING CURVES ARE ALSO SHOWN.)

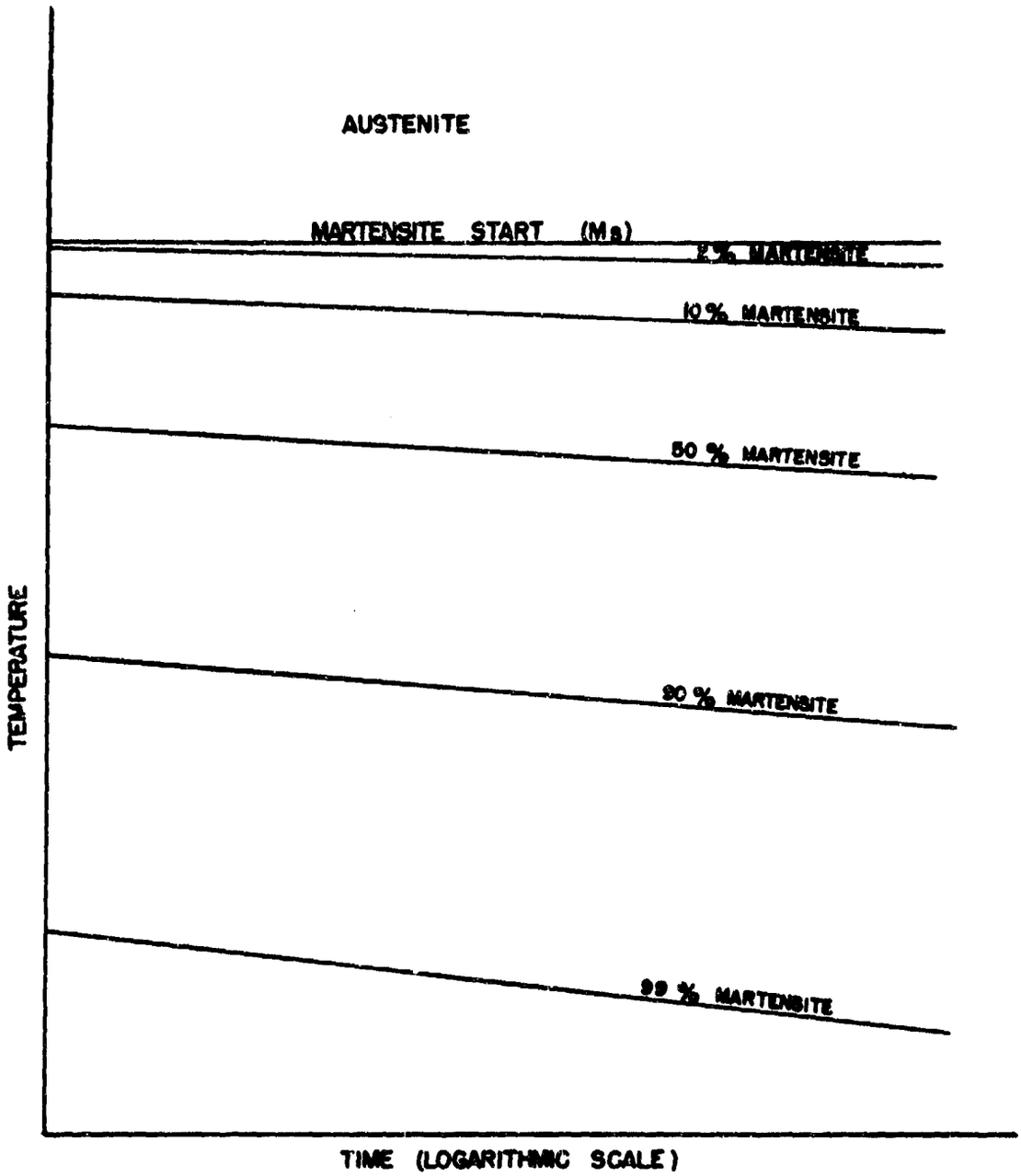


FIGURE 14

AUSTENITE-MARTENSITE TRANSFORMATION ON CONTINUOUS COOLING (SCHEMATIC)

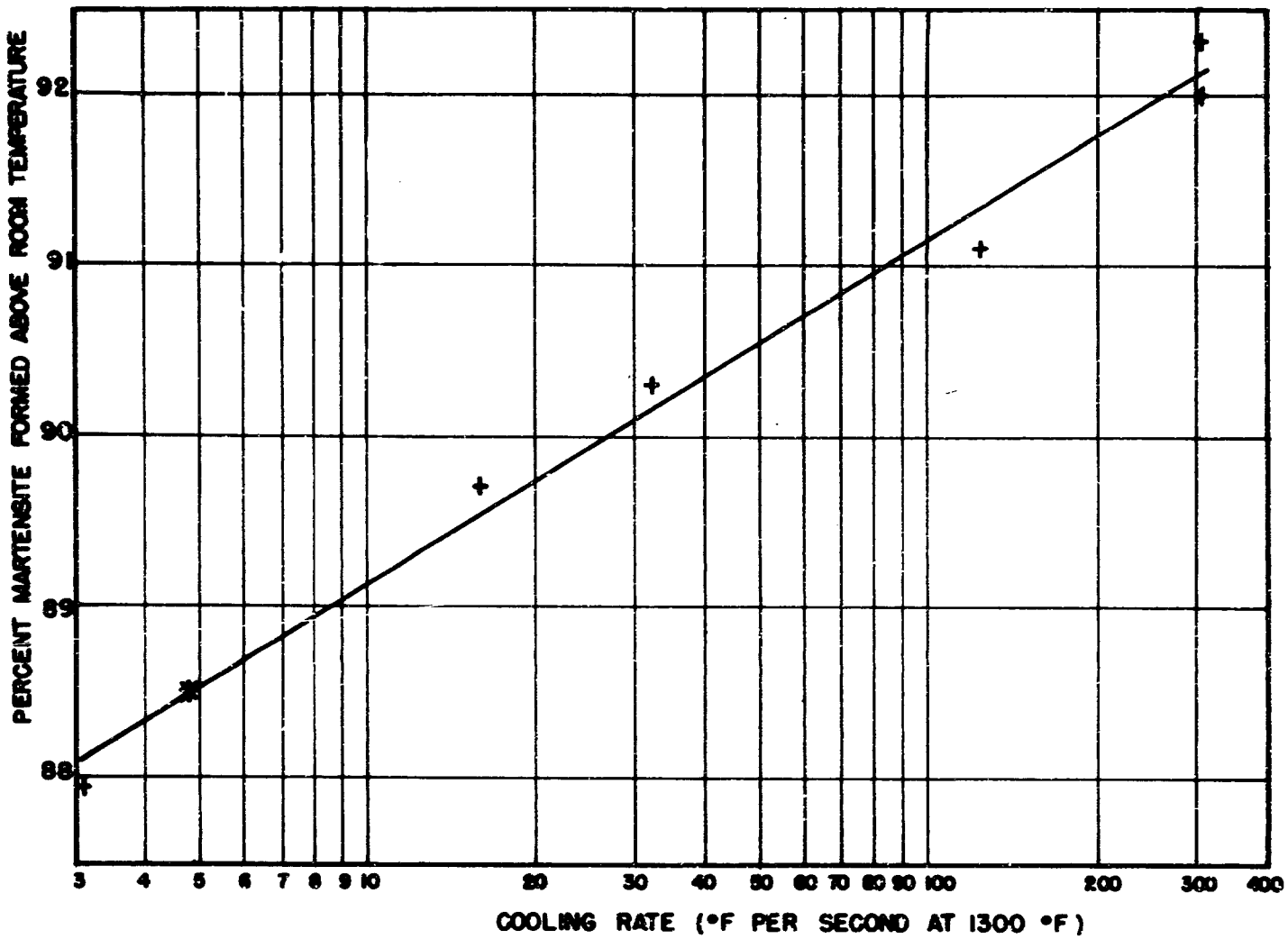


FIGURE 15

EXTENT OF TRANSFORMATION OF AUSTENITE TO MARTENSITE AS A FUNCTION OF COOLING RATE. (.40 % CARBON, MANGANESE-NICKEL-CHROMIUM-MOLYBDENUM STEEL. JOMINY END-QUENCH HARDENABILITY SPECIMEN.)

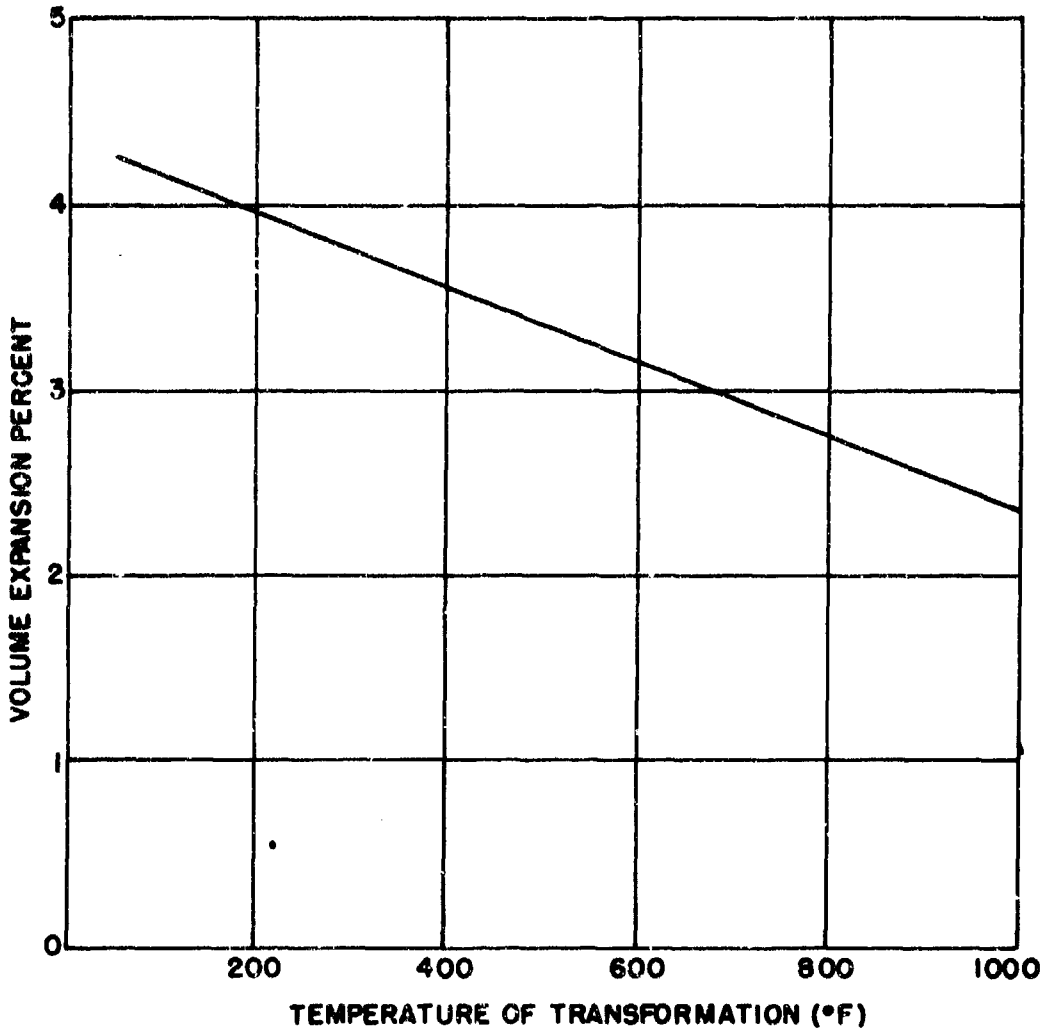


FIGURE 16

EXPANSION ACCOMPANYING TRANSFORMATION OF AUSTENITE TO MARTENSITE

CALCULATED IN PART FROM DATA OF
FLETCHER & COHEN[109]
TAMMANN & BANDEL[110].

FIGURE 17

PHOTOMICROGRAPHS OF MARTENSITE

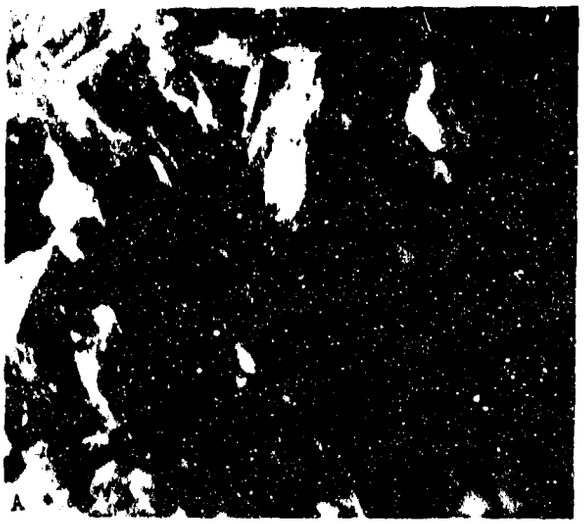
AND TEMPERED MARTENSITE

A. Martensite (X 1000)

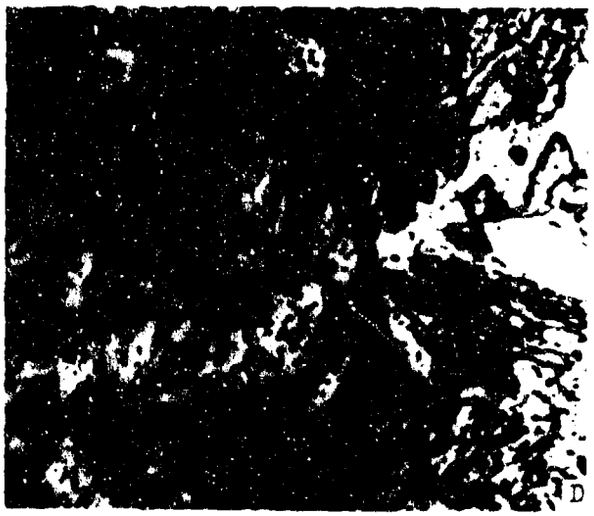
B. Slightly Tempered Martensite (X 2500)

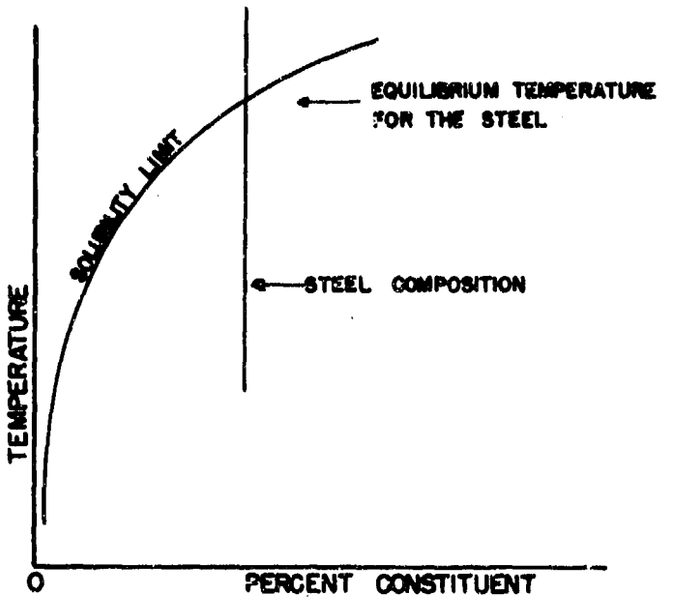
C. Moderately Tempered Martensite (X 2500)

D. Tempered Martensite Spheroids of Carbide in Ferrite Matrix (X 2500)


A

B


D



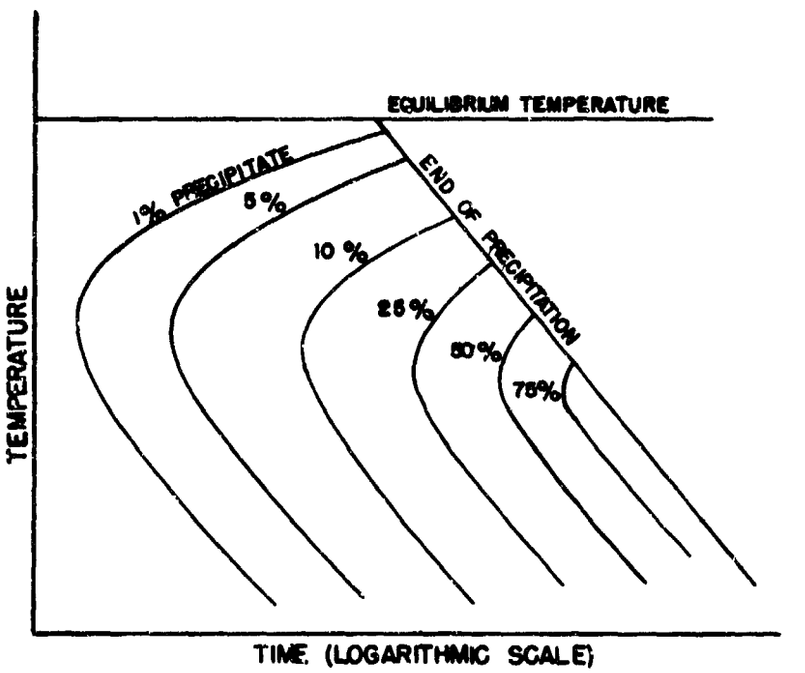


FIGURE 18

TEMPER BRITTLENESS PRECIPITATION (SCHEMATIC)

TOP - EQUILIBRIUM DIAGRAM
BOTTOM - ISOTHERMAL PRECIPITATION DIAGRAM

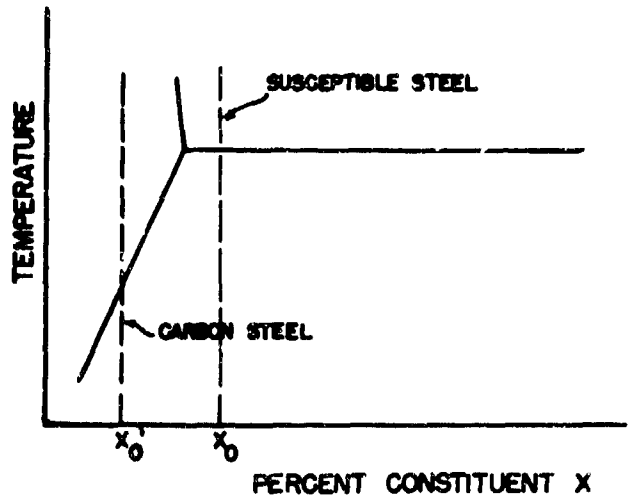


FIGURE 19

POSSIBLE EQUILIBRIUM DIAGRAM
FOR TEMPER BRITTLENESS PRECIPITATION
(SCHEMATIC) (AFTER HOLLOMON)[129]

Suggested Reading

W. Hume-Rothery:
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"Metals", Oxford University Press, London, 1939, pp. 825-1067.

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H. Jolivet:
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J. H. Hollomon:
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II. HEAT FLOW

To obtain in a steel part the transformations desired during hardening and tempering, it is necessary to subject the part to thermal treatments. In such treatments, the periods during which the temperature is changing are as important as the periods during which it is constant. The changes in temperature are governed in general by heat flow in and out of the part. Heat flow within and across the surface of steel parts is, therefore, one of the fundamental phenomena involved in metallurgical design for hardening.

The heat flow may occur during heating or cooling; may or may not be accompanied by phase transformations; may be in a large part or a small, a simple or complex; may involve a gaseous medium, a liquid medium that remains liquid, a liquid medium that boils, freezes, or decomposes at the surface of the metal, a solid medium, no medium at all (a vacuum), or several media. The theory of heat flow is rather well developed. Its fundamentals were established and many of its details investigated at the beginning of the 19th century(130). If the thermal characteristics of the metal and of the metal-medium interface (and certain of the transformation characteristics of the metal) are known, the theory of heat flow permits calculation of the temperature as a function of time at any position in the part, or of whatever other heat flow information may be desired. It is true that the equations involved are difficult to solve, and it is often necessary to employ graphical methods, mathematical machines, or analogy methods for their solutions. The chief difficulty in practice, however, is that the thermal characteristics of the metal and of the interface are usually not known. These characteristics include the thermal conductivity of the metal, its heat capacity per unit volume, and the film coefficient which characterizes the interface. The film coefficient is defined as the ratio of the rate of heat flow across a unit area of interface to the temperature difference across the interface (between the metal surface and the bulk of the medium). If a transformation occurs during the heat flow, the heat evolved during the transformation plays a part, and the occurrence and extent of the transformation as a function of past and present temperature and time, therefore affect the heat flow.

Factors Affecting Heat Flow

Cooling in a Liquid That Boils

As an important illustration of the factors affecting heat flow, let us consider the quenching of a steel part of simple shape, initially consisting of homogeneous austenite at a uniform temperature above the critical range, into a large quantity of a liquid medium that boils considerably below the initial temperature of the part. When the part is immersed in the medium, heat flows across its surface into the adjacent liquid. The temperature of the metal surface decreases, and the temper-

ature of the adjacent liquid rises. Within the metal, heat flows from the hot center of the now cooler surface. (Heat flows from within the part primarily to the nearest surface; little heat flows to other surfaces.) The temperature of the center begins to fall, but remains higher than that of the surface. The temperature distribution within the part assumes the shape shown schematically in Figure 20 (a, b, c). Heat flow through the steel and heat flow across the interface are related both through the temperature of the surface, and through the principle of the conservation of heat, which requires that the rate at which heat reaches the surface from the interior of the part be equal at every instant to the rate at which heat passes across the interface.

Outside the part, the adjacent liquid is very rapidly heated to the boiling point and transformed into vapor. A thin layer of vapor is thus set up around the part, and heat from the part must pass through this layer to reach the liquid. Heat transfer through the vapor is rather poor; that is, the film coefficient is low. Nevertheless, since the difference in temperature between the metal surface and the liquid is high, heat will flow across the interface at a moderate rate and the surface and center will cool, as indicated in Figure 20 (d, e, f). Since the steel is austenitic, the heat flow within the part is governed by the thermal conductivity and the heat capacity (per unit volume) of the austenite. Temperature changes are rapid when the conductivity is high and the heat capacity low. These quantities depend both upon the composition of the austenite and upon its instantaneous temperature.

That portion of the quenching during which the vapor layer exists is frequently referred to as "stage A cooling"*. During this stage the film coefficient of the interface is influenced by the heat of vaporization of the liquid, by the heat capacities and thermal conductivities of liquid and vapor, by the boiling temperature, and probably by the movement of the steel relative to the liquid and by the turbulence within the liquid. These last probably affect the thickness of the vapor layer, rapid movement or turbulence tending to "wash away" vapor mechanically. An increase of any of the factors mentioned will increase the film coefficient. Changes in the heat of vaporization and the boiling point of the medium probably have the greatest effect. (Another factor, temperature of the medium, will be discussed later, because it is somewhat more complex than those mentioned above.)

As the surface temperature falls, the vapor layer becomes thinner, and eventually liquid breaks through in spots and comes in contact with the steel. This contact is only momentary at each spot, since the liquid touching the steel boils quickly. However, as cooling continues, a condition is

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*The existence of several stages of quenching was first suggested by Benedicks(131) in 1904. The presence and nature of the three stages described in this chapter were established in 1920 by Pilling and Lynch(132), whose findings have been confirmed by several other investigators(133-136).

set up in which at each spot there is alternately steel-liquid and steel-vapor contact. This condition is known as "stage B cooling". The surface temperature at which the transition from stage A to stage B takes place is determined primarily by the boiling point of the medium, the interfacial tensions between steel, liquid, and vapor, the relative motion of the part and the medium, and the turbulence of the medium. The higher the boiling temperature, the higher is the temperature at which the vapor film begins to break down. The lower the interfacial tension between steel and liquid as compared with the sum of the interfacial tensions between steel and vapor and between vapor and liquid, the easier it is for a steel-liquid interface to be set up instead of steel-vapor and vapor-liquid interfaces. Motion and turbulence mechanically wash away the vapor layer and permit momentary liquid-to-metal contact.

Since heat transfer takes place much more readily across a simple metal-liquid interface than through a complex one involving transfer from metal to vapor, through the vapor layer, and from vapor to liquid, the film coefficient and the rate of heat flow (insofar as it is controlled by the interface) are much greater in stage B cooling than in stage A. Cooling during stage B is affected by the same factors and in the same direction as cooling during stage A; however, turbulence of the liquid and its movement past the steel play a major role in stage B. The interfacial tension effect mentioned above is an additional factor.

As the temperature of the metal surface approaches and reaches the boiling temperature of the liquid, boiling ceases and Stage C cooling begins. Here the gas phase is not involved, and heat flowing across the interface serves only to heat the nearby liquid, not to vaporize it. This heat is then distributed to the rest of the liquid by conduction and convection. Factors affecting the film coefficient in stage C include the heat capacity and thermal conductivity of the medium, the relative movement of medium and steel, and the turbulence of the medium.

The rate of the flow of heat per unit area across the interface is equal to the product of the film coefficient and the temperature difference between the metal surface and the bulk of the liquid. As quenching proceeds, the temperature difference decreases, and a decrease of heat flow across the interface is, therefore, superimposed upon the effects due to changes in film coefficient. Since the temperature difference approaches zero as the temperature of the surface approaches that of the liquid, the heat flow also approaches zero. On this basis, it will take an infinite time for the temperature of the part to reach that of its surroundings, but the difference between part and liquid temperatures continuously decreases.

The effects of temperature of the liquid upon flow of heat across the interface have not yet been discussed; they are of two kinds. In the first place, increasing the temperature of the liquid decreases the temperature difference between the metal surface and the liquid. The effect of this decrease upon the rate of heat-flow is most marked when the temperature difference is small, as towards the end of a quench. In the second place,

changes in temperature of the liquid affect the film coefficient in several ways. Thus, increases in temperature decrease the viscosity of the medium. This decrease in viscosity increases convection and turbulence, tending to increase the film coefficient during all stages and to increase the temperature of transition from stage A to stage B. Increase in medium temperature decreases the additional rise in temperature necessary to bring the liquid to its boiling point. This may markedly lower the temperature of transition from stage A to stage B cooling. It also tends to decrease the film coefficient during stages A and B.

Not only the surface conditions (film coefficient) but also the thermal conductivity and heat capacity per unit volume of the metal can be expected to change during cooling. The thermal conductivity of austenite increases considerably as the temperature is lowered, while the heat capacity decreases somewhat. Since the metal near the surface is colder at any instant than the metal near the center, the conductivity and heat capacity will not be constant across the section. Moreover, at some temperature the austenite will start to transform to martensite. Thereafter, the pertinent conductivity and heat capacity will not be those of austenite but those of mixtures of austenite and martensite in varying percentages, and finally, perhaps, of martensite alone. During the austenite-martensite transformation a considerable evolution of heat occurs. Since the progress of this particular transformation, and hence the amount of heat liberated per degree cooling, is only slightly dependent upon rate of cooling, it is probably possible to consider the heat evolution as simply an increase in heat capacity. Again, however, the heat capacity will vary across the section, since the temperature does.

If proeutectoid, pearlite, or bainite reactions occur during cooling, the situation becomes more complex. The occurrence and progress of these reactions depend not only upon the transformation characteristics of the austenite as fixed by its composition, grain size, and homogeneity, but also upon the temperature-time relations during cooling. Transformation from austenite to these other constituents involves changes in thermal conductivity and heat capacity, as well as heat evolutions that may be sufficient to raise the local temperature despite the heat loss across the interface. (This phenomenon is known as "recalescence".) Such heat evolutions cannot be considered as simply changes in heat capacity, but must be taken into account separately. Thus the instantaneous conductivity, heat capacity, and rate of heat evolution at any point depend upon the temperature at each point throughout the part as a function of time up to the instant under consideration. Not only austenite decomposition, but tempering of the resulting decomposition products may occur during quenching. Heat flow in a quenched part is, therefore, to some extent dependent upon the transformation characteristics of the steel, just as the occurrence and progress of the transformations are, to a greater extent, dependent upon the heat flow.

Cooling in Media of Various Types

When a steel part is quenched in a liquid medium that boils or decomposes at a temperature below the initial surface temperature of the part, the behavior during quenching is ordinarily that described above. Differences between the various media that may be used lie only in the numerical values of the various factors governing the film coefficient. Thus, dilute aqueous solutions seem to differ from water appreciably only in the temperature of transition from first to second stage quenching. Solutes such as sodium chloride and sodium hydroxide raise this temperature while soaps apparently lower it (132, 134, 135, 137). This behavior may perhaps be associated with the effect of the solutes in raising or lowering, respectively, the interfacial tension between the liquid and the steam while having less effect upon the interfacial tension between liquid and metal and between metal and steam. When oils are used as quenching media the constants involved are likely to be considerably different from those for water, but the three characteristic stages are present in both cases. Oils may decompose as well as boil in the early part of the quenching, but as far as heat flow is concerned this simply changes the numerical values of certain factors. Some molten salts may decompose into both gaseous and liquid products without boiling; the gaseous decomposition products affect the heat flow in a manner similar to vapor. (However, while vapor may condense to the original liquid, decomposition products may not condense or recombine and so may be removed only by convection.) The temperatures of transition from A to B stage for molten salts, as well as some other media, are likely to be above the initial temperature of the steel, so that stage A cooling will be absent. The temperatures of transition from the B to C stage for molten lead and some molten salts may be above the initial temperature of the steel, so that only stage C cooling will be present. With any medium, lowering the initial temperature of the steel makes it more likely that the A and the B stages will be absent.

In air or gas cooling, only stage C can be present. Heat flow by conduction radiation is more important in air or gas cooling than in liquid cooling. Radiative heat transfer will occur when the medium is not opaque to the visible and infra-red wave-lengths emitted by steel. When the radiation is absorbed in the bulk of the medium, cooling can be considered to be stage C, but the factors affecting the film coefficient are somewhat different than when cooling takes place by conduction or convection. The heat capacity, conductivity, motion, and turbulence of the medium will have only a slight effect.* (They may still have some effect since they affect dissipation of heat in the medium after the radiation is absorbed.) The emissivity of the metal surface is, however, the most important factor during radiation: the film coefficient will vary directly as the emissivity. Moreover, the film coefficient will vary, to a first approximation, as the cube of the absolute temperature of the surface.

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*These factors have little effect on the radiative heat flow in stage C cooling but the heat flow by convection will be affected.

Since the emissivity of a surface is equal to one minus its reflectivity, a clean, highly reflective steel surface will have considerably lower emissivity and lower film coefficient for radiation than a dark, oxidized ("scaled") surface. Plating and other surface conditions will also affect the emissivity. Such changes in emissivity may be important not only in stage C cooling, but also in stage A, since radiation may play a considerable part in heat transfer across the vapor layer. Radiative heat transfer across vapor bubbles may also occur during the B stage of liquid quenching.

By changing the effective interfacial tensions and possibly, through friction, the turbulence, surface conditions can also affect the film coefficient during stages A and B, as well as the temperature of transition from A to B. If the plating, or especially scale, is thick enough, its thermal conductivity and possibly even its heat capacity may affect heat flow. Since conductivity and heat capacity are properties of the part and not of the interface, they may be of importance regardless of the stage of cooling.

When radiation passes through a medium essentially transparent to it, and is absorbed not in the medium but in surrounding walls, the important temperature difference is not that between metal surface and medium but between metal surface and wall. If the film coefficient concept is to be used, it must refer to the metal-wall "interface", rather than the metal-medium interface. The film coefficient will be greater the higher the emissivity of the part and the lower the emissivity of the wall. Unless the part is entirely surrounded by walls of uniform temperature and emissivity, geometrical considerations enter: The film coefficient at any point on the part is affected by the solid angle of wall "visible" from the point and its temperature and emissivity. Specular reflection characteristics and hence the angles between wall and part surfaces may play a part. If radiation from the part changes the wall surface temperature appreciably, the thermal conductivity and heat capacity of the wall enter.

This last case approaches that of a solid cooling medium separated from the part by a thin layer of gas. Heat flows across the gas layer by conduction, convection, and radiation; through the solid medium by conduction. If the heat flow changes the surface temperature of the solid medium by only a small amount relative to the temperature difference between part surface and solid medium surface, the situation does not differ essentially from C stage cooling in a gaseous or liquid medium. If, however, the temperature change of the solid medium surface produced by the heat flow is appreciable compared to the difference between the temperatures of the two surfaces, not only must the film coefficient and the thermal conductivity and heat capacity of the steel be considered, but also the conductivity and heat capacity of the solid medium. If the characteristics of the solid medium are such that an appreciable portion of the heat from the part reaches the outer boundary of the medium before the part is essentially down to its final temperature, the conditions at this outer boundary as well as the dimensions of the solid medium may affect the heat flow also.

The cooling medium may consist of a mixture of several phases, but the heat flow under such circumstances usually differs only in numerical values from the flow with a single phase medium. (Cooling in ashes, a solid-gas mixture, is an example; such cooling will differ little from cooling in a solid.) Several cooling media may be effective at the same spot. (For example, a surface may be simultaneously cooled by stage C cooling to the adjacent air and by radiation to the walls.) Different cooling media may be applied to different portions of the same part. (One end of a bar may be quenched by water while the other end cools in air.) Even if the same medium is applied to the whole part, different portions of the surface may, at the same moment, be in different stages of cooling, because of differences in local surface temperature or in medium turbulence. (This is common in water-quenching a part having sections of widely differing thicknesses.)

Heating

Heat flow on heating is simply the reverse of that on cooling. Neither boiling nor decomposition of a liquid medium occurs when a cold part is quenched into it, but a gaseous medium may condense on the surface of the part, giving rise to a condition, corresponding to A stage cooling, that may be termed A stage heating. Since turbulence in the gas would be unlikely to remove the liquid wholly even from local areas of the surface, no condition corresponding to B stage cooling is likely to arise. Instead, as the metal surface approaches the boiling point of the medium, a transition from A to C stage heating will occur. A stage heating may also occur when a part is heated in a liquid which solidifies at the surface of the part.

C stage heating, by conduction, convection, or radiation from a medium that does not condense or solidify, does not differ in any important respect from C stage cooling. Heat flow directly from the walls to the part can also occur, by radiation. In the case of radiation it should be remembered that the film coefficient varies approximately as the cube of the absolute temperature of the hotter body, which in heating is not the part but the medium or walls.

Transformations involving heat absorptions, such as the formation of austenite, may occur on heating. Such heat absorption corresponds to heat evolution on cooling.

Equalization

Of some metallurgical interest is temperature equalization throughout a part in which the temperature is initially non-uniform because of prior heating or cooling. If there is no heat flow across the part surface all portions of the part will approach a uniform temperature. Heat flow within the part under such conditions is affected by the size and shape of the part and by its thermal conductivity and heat capacity per unit volume. (If transformations occur, heat evolution or absorption

will have an effect.) The initial condition here is not one of uniform temperature throughout the part, but of a known initial temperature distribution. (In some important practical cases of cooling or heating also, the initial temperature of the part is not uniform.)

Temperature Distribution During Heat Flow

During cooling, the temperature distribution within the part will successively assume a series of shapes such as those shown in Figure 20 (b, e, and h) (provided no heat evolutions take place). On heating, the distribution will assume the shapes given by these same sketches turned upside down. Most of the temperature gradient will be across the interface if the film coefficient (M) of the interface is low relative to the ratio of the thermal conductivity (K) of the metal to the distance (D/2) the heat must flow from center to surface of the part $\left(M \ll \frac{K}{D/2} = \frac{2K}{D}\right)$.

The temperature distribution then will assume the form indicated in Figure 20 (a, d, and g), where the temperature across the section of the part is relatively uniform. Most of the temperature gradient will be within the part if the film coefficient is high relative to the ratio of the thermal conductivity to the distance the heat must flow within the part $\left(M \gg \frac{2K}{D}\right)$.

The temperature distribution then will assume the form indicated in Figure 20 (c, f, and i). Lowering the ratio of thermal conductivity to heat capacity (per unit volume) and increasing the distance the heat must flow within the part both tend to increase the temperature differences within the part, relative to the temperature difference across the interface.

Simplification of Heat Flow

The discussion above gives some indication of the multiplicity of factors affecting heat flow during heat-treatment of steel parts. To consider their individual effects quantitatively and in detail is beyond the scope of this book, though the effects of a few factors will be reviewed qualitatively in the subsequent chapter, "Quenching". However, (neglecting such heat evolutions and absorptions as cannot be considered as changes in specific heat,) the heat flow for the great majority of cases that are of interest in heat-treatment is governed by the initial conditions, the thermal conductivity and heat capacity (per unit volume) of the metal, the film coefficient of the metal-medium interface, and the size and shape of the part. As has been previously indicated, the numerous factors affecting heat flow do so through their effect upon these relatively few characteristics.

The initial conditions consist of the initial temperature at all points in the part and the initial temperature of the medium. The temperature of the medium ordinarily can be considered to be uniform and constant. The thermal conductivity and heat capacity (per unit volume) of the

metal depend on composition, microstructure, and local temperature. (If transformations may occur during the heat flow, so affecting the microstructure, the transformation characteristics as a function of time and temperature influence the conductivity and specific heat.) The film coefficient is affected by many variables, but in most cases the heat flow can be divided into three or fewer regions of surface temperature in each of which the film coefficient is very nearly constant. The size that is of primary importance is the thickness of the part. The shape, in a great many cases, can be considered as a combination of simple shapes such as plates, cylinders, and spheres.

Considerable information can be obtained from a mathematical treatment of even the simplest cases, such as that in which the initial temperature of the part is uniform, the thermal conductivity, heat capacity (per unit volume) and film coefficient are constant during the heat flow, and the shape is a plate or a right circular cylinder*. Information of greater applicability could be obtained by considering the film coefficient to have one of three successive values, depending on the surface temperature (the value for the B stage being higher than for the other two), but no treatment of the three-stage case seems to have been made as yet.

For the one stage case, under the conditions mentioned above, it is convenient to specify the interface characteristics in terms of the severity of quench, H, which is defined as the ratio of the film coefficient to twice the thermal conductivity of the metal ($H = M/2K$). It is also convenient to specify the heat flow characteristics of the metal in terms of the thermal diffusivity, a, which is defined as the ratio of the thermal conductivity to the heat capacity per unit volume. Plates with the same value of the product HD, where D is the thickness, will pass through the same temperature distributions during heat flow, as indicated in Figure 20. The same is true for cylinders with the same value of the product HD. The time after the start of the heat flow at which parts reach these distributions is directly proportional to the square of their thickness and inversely proportional to their thermal diffusivity. For such comparisons, temperature is conveniently expressed as a fractional temperature, U: the difference between the instantaneous and the medium temperatures divided by the difference between the initial and the medium temperatures.

Equations and tables (138-142) for the fractional temperature U at any point within a plate or cylinder, at any time, have been developed for the conditions mentioned in the preceding paragraphs, and are based upon the product HD. These equations and tables are very useful in calculating heat flow conditions during heat-treatment, despite the special conditions assumed in their derivation. The most extensive tables published appear to be those of Russell(141), which have been reprinted in Austin's book(143).

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*The assumed constancy of the film coefficient is often referred to as "Newton's Law of Cooling".

As indicated in Figure 21, high HD values result in large temperature differences within the part, while low HD values produce small temperature differences. For the limiting case of low HD values, the temperature across the part is almost uniform, and the time-temperature curve is very close to an exponential function, in which the rate of change of the fractional temperature is proportional to the fractional temperature:

$$-dU/dt = cU \qquad (2\text{-}1)$$

where t is the time and c a constant. Then:

$$\ln U = -ct \text{ and } U = e^{-ct} \qquad (2\text{-}2)$$

where e is the base of the natural logarithm.

The value of c is:

for plates: $HD(\frac{4a}{D^2})$

for cylinders: $2HD(\frac{4a}{D^2})$

for spheres: $3HD(\frac{4a}{D^2})$

where a is the thermal diffusivity.

Equation (2-2) holds within a few percent for values of HD smaller than 0.1*. For larger values of HD there is an appreciable initial transient, during which the center cools or heats more slowly and the outside more rapidly than Equation (2-2) would indicate. (Figure 21, top.) However, as the transient dies away, the rate of cooling or heating approaches that given by Equation (2-1). The fractional temperature then approaches the relation:

$$U = e^{-c(t + t_0)} \qquad (2\text{-}3)$$

where t_0 is a constant, positive for positions near the surface and negative for positions near the center of the section.

Equation (2-2) gives the surface and center temperatures accurately only for low HD values. At higher HD values, the deviation of the surface temperature from that indicated by the equation is in the opposite direction from the deviation of the center temperature. Equation (2-2) therefore provides, up to moderate values of the product HD, a fair approximation to the average temperature of the section. The error introduced by this assumption is such that the equation gives higher value of

*HD is dimensionless; the value of H must be expressed in units the reciprocal of those in which D is expressed.

the average U than a more exact calculation, because equation (2-2) essentially assumes that heat flow is limited only by interface conditions, whereas it is in fact limited also by conduction within the part.

The preceding discussion has been concerned with simple shapes, such as plates and cylinders. More complicated shapes can generally be considered as combinations of simple shapes. Thus, a long rectangular bar can be considered as the intersection of two perpendicular plates. At any time, the value of the fractional temperature U for a point within the bar will be the product of the value of U for a point at the same position within two plates whose thicknesses are equal to the two shorter dimensions of the bar. (The value of HD, it must be remembered, will ordinarily be different for the two plates if their thicknesses are unequal.) A cube can similarly be considered as the intersection of three perpendicular plates of equal thickness: U at the cube center, for example, will be equal to the cube of U for one of plates at the same instant. A short round bar can be considered as the intersection of a cylinder with a plate perpendicular to the cylinder axis.

Computations made in this way indicate that, if the value of H is constant over the surface, the heat flow is governed primarily by the smallest dimension. Little error is involved, for example, in considering a round bar to be a cylinder unless the length is almost as short as the diameter, or in considering a slab of finite length and width as an infinite plate unless the width is almost as small as twice the thickness (144). As another example, a threaded round can be considered as an unthreaded round of slightly more than the minor diameter.

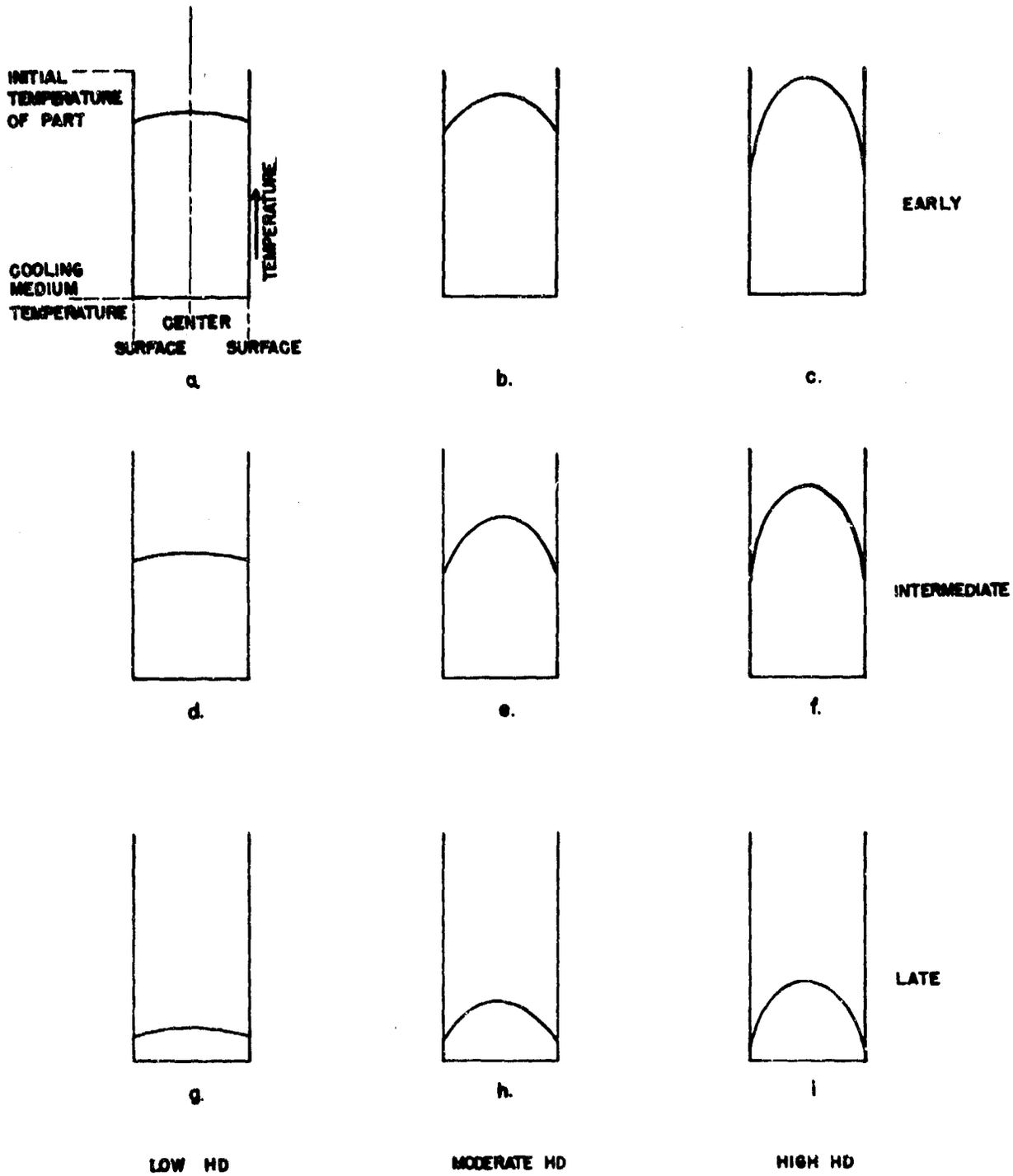
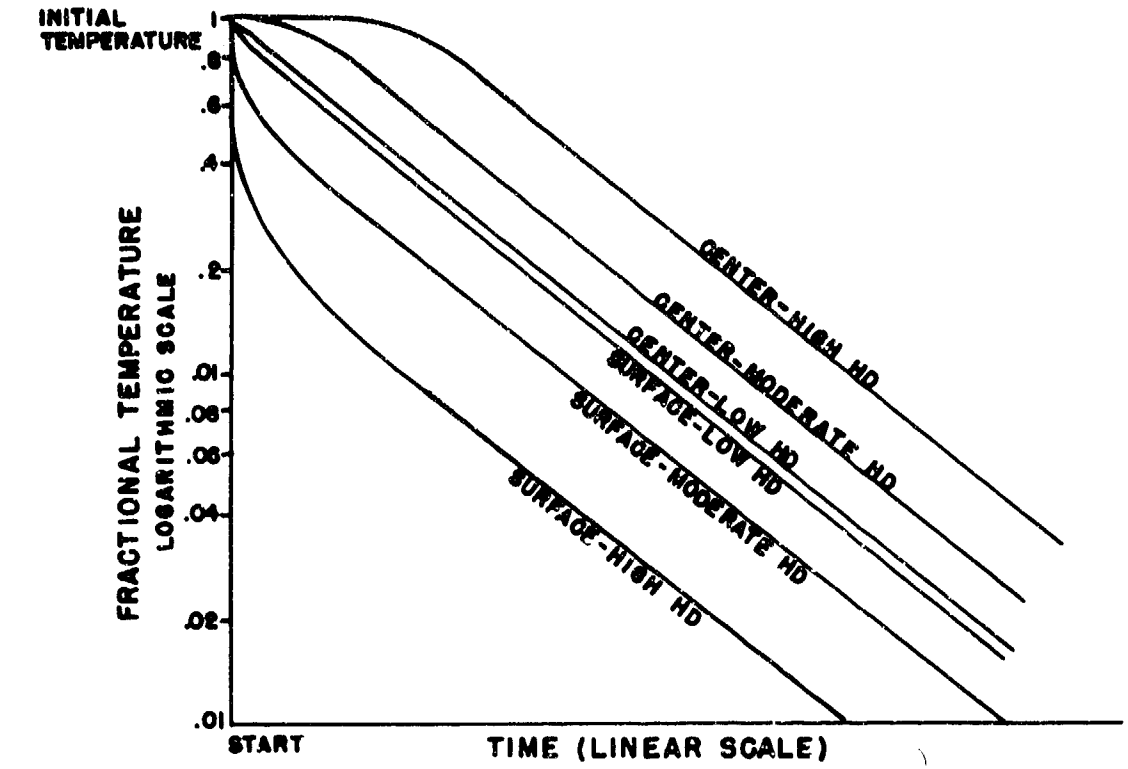


FIGURE 20

EFFECT OF HD PRODUCT UPON TEMPERATURE DISTRIBUTION ACROSS A SECTION DURING QUENCHING. (SCHEMATIC)



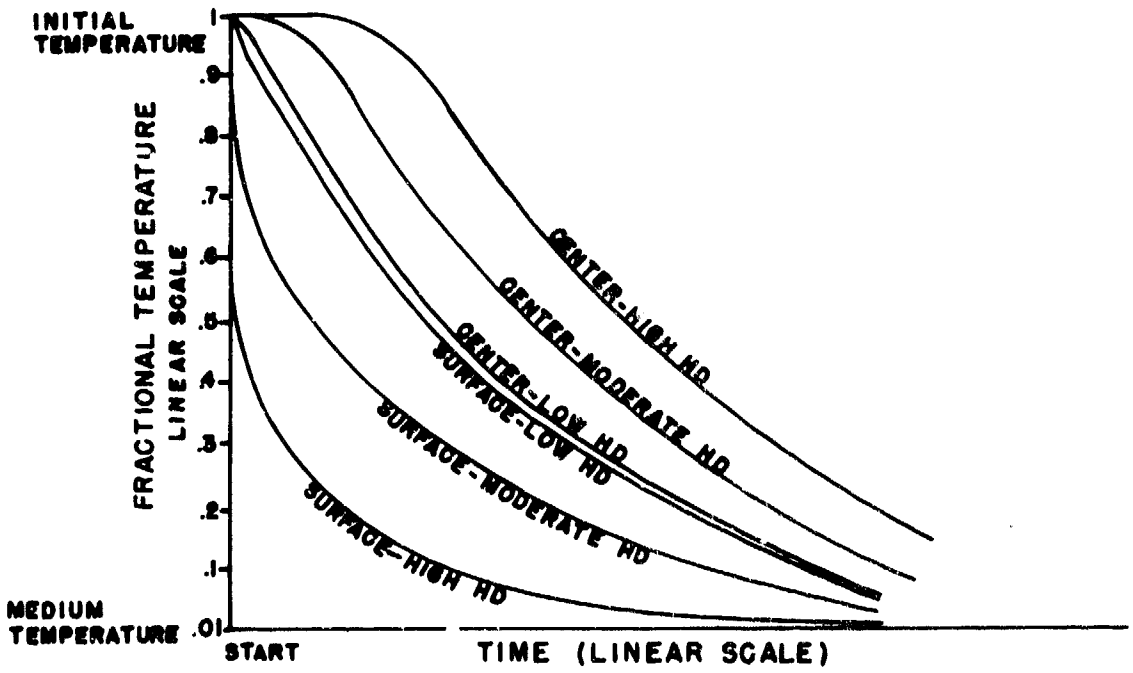


FIGURE 21

COOLING OR HEATING CURVES FOR VARIOUS HD VALUES.
(SCHEMATIC)

TOP - LOGARITHMIC TEMPERATURE SCALE
BOTTOM - LINEAR TEMPERATURE SCALE

Suggested Reading

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III. MECHANICAL BEHAVIOR

The object of the study of the mechanical behavior of metals is to relate the effects of changes in structure to changes in mechanical properties. If an attempt were to be made to establish individual correlations between all the mechanical properties and structure, the amount of experimental data necessary would be tremendous. A better approach is to study the fundamentals of mechanical behavior that underlie the properties, and the effects of metallurgical variables (micro-structure) upon this behavior. Once these fundamentals are understood, the relations between the metallurgical variables and the various mechanical properties can be found simply.

Stress-Strain Relations

Although the mechanical behavior of metals is determined by certain phenomena that take place on a microscopic or submicroscopic scale (slip, twinning, local viscous flow, etc.), a discussion of these phenomena would be beyond the scope of this volume*. The present discussion will be confined to behavior on a macroscopic scale. The macroscopic phenomena that govern the ordinarily-measured mechanical properties of metals are plastic flow and fracture. The simplest and most fundamental measures of the flow and fracture characteristics are stress-strain curves. A single stress-strain curve does not by any means determine the entire mechanical behavior, but an understanding of the significance of such curves leads to a consistent interpretation of the results of standard mechanical tests, and of the effects of metallurgical structure upon these results.

As a load is first applied to a specimen of a polycrystalline metal, the stress is proportional to the deformation (change in dimensions), and the metal is said to be elastic. As the stress is increased, plastic flow begins and the metal strain-hardens; plastic flow increases the stress required for further plastic flow. Strain-hardening continues until the specimen fractures. In general, the strain-hardening characteristics of metals are measured by determining the stress-strain curves of specimens deformed in uniaxial tension. In this case, the stress is the load divided by the area of the specimen at that load. The change in strain is defined by the following relation:

$$d\varepsilon = \frac{dl}{l} \qquad (3\text{-}1)$$

where ε is the strain and l the length over which the change in length dl occurs. The strain, ε, is found by integration to be:

$$\varepsilon = \ln l/l_o \qquad (3\text{-}2)$$

*A review of these phenomena may be found in the book, "The Structure of Metals" (145).

where l_o and l are the original and the instantaneous lengths respectively (146, 147). During plastic deformation the volume of the metal is essentially constant* and therefore:

$$A_o \cdot l_o = A \cdot l \qquad (3\text{-}3)$$

where A_o and A are the original and instantaneous areas respectively. Upon substitution in Equation (3-2) the following relation(148, 149) results:

$$\epsilon = \ln \frac{A_o}{A} \qquad (3\text{-}4)$$

The instantaneous diameter can be measured and the strain obtained from Equation (3-4).

If the stress (load divided by the instantaneous area) is plotted as a function of the strain as defined above, schematic stress-strain curves of the type presented in Figures 22 and 23 are obtained for most metals. For steels, the curves consist of three rather distinct regions: the elastic, the initial yielding, and the plastic flow. Duri g elastic deformation (from origin to Y in Figure 22, to Y' in Figure 23) the stress is proportional to the strain; if the load is removed the strain returns to zero. The constant of proportionality between tensile stress and strain is referred to as the elastic modulus (or as Young's modulus) and for steel is equal to approximately 30,000,000 p.s.i. at room temperature. The elastic modulus decreases slightly as the temperature increases, but for steels of the moderate alloy contents generally considered for heat-treated parts it is essentially independent of all other variables, including microstructure.

At Y and Y' in Figures 22 and 23 yielding occurs and the curves deviate noticeably from straight lines. Yielding is the initiation of plastic deformation, but because there are certain pecularities (Y' to A, Figure 23) sometimes found in the initial stages of plastic flow that are not found later, it is often convenient to speak of initial yielding separately from later flow.

After the initial yielding, homogeneous plastic deformation occurs (Y to F and A to F). If the load is removed the specimen is found to be permanently deformed. The stress-strain curve in this region deviates from a straight line; it rises continuously. Plastic deformation increases the stress required for further plastic deformation and the metal strain-hardens. The slope of the stress-strain curve at each value of the strain represents the rate of strain-hardening for the particular metal deformed a specified amount (in tension). As deformation proceeds the rate of strain-hardening decreases.

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*Except at small strains, the change in diameter due to the change in volume accompanying the elastic deformation will be small compared to the change in diameter arising from plastic flow.

If some plastic deformation has occurred, and the stress is removed, the plastic deformation ceases and the elastic deformation disappears. However, some additional anelastic deformation remains and disappears only gradually, over a period of seconds or hours. Similarly, if the load is being increased, the deformation does not correspond to the instantaneous stress. Rather, if a load is reapplied after plastic deformation, the strain assumes a given value almost instantaneously and then further deformation occurs as the load is maintained. This additional deformation is not plastic in the ordinary sense, for if the load is removed, the strain will be found not to be permanent. Furthermore, it is not elastic in the strict sense since the strain does not reach the final value immediately. Thus the deformation may be called anelastic. This anelastic deformation is of importance to the understanding of the fundamentals of plastic flow and is related to the elastic aftereffect and Bauschinger effect discussed later. The details of this type of deformation discussed elsewhere(103) will not be further considered here.

At F in Figures 22 and 23, fracture occurs. However, since plastic deformation has taken place the metal that fractures at F is different from the one that yielded at Y or Y'. The fracture stress at the strain ϵ_f applies only to a metal deformed to this strain in simple tension.

The tensile stress-strain curve may be considered as a locus of points representing the tensile stress required for plastic flow of an infinite number of specimens of the same prior metallurgical structure but each differing from every other specimen in that it has been deformed a different amount in simple tension. The specimen deformed to the tensile strain ϵ_f (Figures 22 and 23) fractures and does not flow. The stress value for this specimen falls not only on the locus of flow stresses but also on another locus of stresses(150): that of the fracture stresses. It may be postulated that for each of the specimens deformed to strains less than ϵ_f, a fracture stress exists which is greater than the stress required for plastic flow, for the specimen flows and does not fracture*. Thus a locus of fracture stresses can be conceived which intersects the locus of flow stresses at the strain ϵ_f (Figure 22). The curves will be referred to as fracture-stress and flow-stress curves respectively. The concept that there exists a fracture-stress curve in addition to the ordinary flow curve affords a means of interpreting, in a consistent manner, the effects of structure on the mechanical properties. Not only may a change in metallurgical structure alter the flow curve but it can and does induce tremendous changes in the fracture stress curve.

Effect of Metallurgical Variables

The effects of metallurgical variables on the mechanical properties may be interpreted in terms of their effects upon the flow and fracture curves. Most of the studies of the effects of metallurgical

*A method of obtaining an estimate of the locus of fracture stress is discussed on page 56 and following.

variables on flow and fracture have been made by performing tensile tests. Tensile tests designed to determine the effect of metallurgical variables are generally performed at room temperature (about 70° F., 20° C.) and at moderate rates of strain (about 10^{-4} sec.$^{-1}$). In the following paragraphs the effects of metallurgical variables upon the tensile stress-strain curves obtained for steel in such normal room-temperature tensile tests are discussed. The discussion is divided into three parts; the first two deal with the effects of variables upon the initial yielding and upon the later portion of the flow stress curve, while the third deals with their effects upon the fracture stress curve.

Yielding

The type of initial yielding is highly dependent upon metallurgical structure. For some steels the stress-strain curves deviate only gradually from linearity, as illustrated in Figure 22. For others, the deviation is abrupt, as shown in Figure 23. For steels of the latter type the deformation originates in discrete bands at stress concentrations and proceeds by the propagation of these bands along the specimen, as illustrated in Figure 24, rather than by uniform extension. This type of deformation has been referred to as the Piobert(152) effect and the discrete regions of deformation are known as Lüders'(153) bands or Hartmann(154) lines. The stress-strain curve associated with steels yielding in this manner is characterized by a lower yield point, B, and an upper yield point, Y'(Figure 23). The magnitude of the upper yield point depends upon the magnitude of the stress concentrations present in the specimen. For example, the more gentle the fillets at the shoulder of the specimen and the more finely polished its surface(155, 156), the more difficult is the formation of Lüders' bands and the higher will be the upper yield stress. The lower yield stress, B, and the amount of non-uniform deformation (BA) depend upon the structure of the steel. Changes in steel structure that increase the lower yield stress without affecting the rest of the flow stress curve will increase the amount of non-uniform deformation (lower yield strain), as illustrated in Figure 25.

It has been shown that the presence of Lüders' type of deformation (as well as its return upon aging, following plastic deformation) is connected with the presence of nitrogen and carbon in the steels(157)*. Less than .002% of either of these elements in iron can cause heterogeneous yielding, presumably because of the precipitation of iron carbides or nitrides at the grain boundaries. Several investigators(158-160) have pointed out how such a precipitate might result in discontinuous yielding. They have suggested that the presence of a grain boundary precipitate restricts the relative motion of grains, causing this restriction rather than the resistance of the grains themselves to limiting the stress re-

*An excellent review of the information available concerning this type of yielding is to be found in the appendix of the paper by Gensamer and Low (157).

quired for plastic flow. Many observations lend credence to such arguments(161). For example, grain size has a marked influence on the magnitude and extent of the lower yield strength(162), arising supposedly from the larger amount of restrain of the grains of fine-grained steels to relative motion. For steels of the same carbon and nitrogen contents, the presence and the extent of the discontinuous type of yielding depend also on the gross microstructure of the steel. As the strength* is increased by dispersion of the carbides, the relative resistance to plastic flow of the grains themselves becomes greater than the resistance to flow imposed by the resistance to relative motion of the grains, and the steel tends less and less to exhibit the drop in load at yielding(163). It appears also that at a given strength level, the more random the distribution of the carbides and the more nearly spheroidal they are, the more likely is heterogeneous yielding(164) to occur. Thus, steels having tempered martensitic structures exhibit the Piobert effect at higher strength levels than the same steels having pearlitic structures(163).

Flow

The height of the stress-strain curve, as measured by the stress required to produce a given small strain (larger than the lower yield strain) appears to depend primarily on the mean free path in the ferrite (164), which in turn is determined by the size, shape and distribution of the carbide particles. The relation between the height of the flow curve and the mean ferrite path is illustrated in Figure 26. Thus, for tempered martensitic steels, the finer and the more numerous the carbide particles, the greater is the strength, and for pearlitic steels, the smaller the separation of the carbide plates the greater is the strength.

Based upon the above analysis, steels of the same strength (stress at a given small strain) should have the same mean ferrite path. A question then arises as to whether the shape of the entire stress-strain curve depends only upon this parameter of the structure or if it depends also upon the shape and distribution of the carbide particles or upon composition.

It has been shown(163) that from very small strains to strains of about .4 the stress required for flow is related to strain in the following way:

$$\sigma = K (\varepsilon)^m \qquad (3\text{-}5)$$

where σ is the stress, ε, the strain, K, a constant, and m a fractional power that depends only upon carbon content and strength (and to a minor

*The word "strength" is used to designate the level of the stress-strain curve as measured by the stress required to produce some small specified plastic strain. It refers to that part of the flow curve beyond the region of initial yielding.

extent on alloy content). Typical curves in which the stress is plotted as a function of the strain on a logarithmic scale are presented in Figure 27 for an alloy steel of several strength levels and having a tempered martensitic structure. For the specimens yielding inhomogeneously the power relation begins only after the strain becomes greater than the lower yield strain.

In Figure 28 the variation of the exponent m (the strain-hardening exponent) with strength and carbon content is illustrated for a series of plain carbon steels. Even though the steels had a wide variety of metallurgical structures, the strain-hardening exponent appears to depend only upon the carbon content and strength level*. The strain-hardening exponents for alloy steels in general appear to be slightly higher than for carbon steels(163). Insufficient data are available, however, to establish any definite relations among the small effects of individual alloying elements. Thus, to a first approximation, strain-hardening depends only upon the strength (which in turn depends only upon the mean ferrite path). Carbon content has some effect and alloy content a minor effect in increasing the strain-hardening at a given strength level.

As has already been mentioned, the relation between stress and strain given by Equation (3-5) extends only to strains of about .4. It has been found that for steels the stress is approximately a linear function of the strain from about the strain at the maximum load (T of Figures 22 and 23) to the fracture strain(149). The slope of this part of the stress-strain curve is referred to as the strain-hardening modulus. This linear relation between stress and strain may be purely coincidental** resulting from the modification of the power relation by orientation effects that depend upon the nature of the deformation(163). This effect will be considered in detail later. Even so, it is frequently desirable to know the relations between stress and strain in this range of strains. These relations are given in Figures 29 and 30. In Figure 29 the strain-hardening modulus is plotted as a function of the stress intercept (I of Figure 22) and in Figure 30, the maximum and minimum moduli obtained by heat treatment are plotted as functions of carbon content.

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*This relation applies only to tempered steels. There is reason to believe that steels having an untempered martensitic structure behave differently.

**It must be remembered also that after necking begins the stresses are no longer uniform in the necked region and the meaning of graphs of average stress versus strain is not precise. (See discussion of Tensile and Related Tests in Chapter IV.)

Fracture

Just as the flow stress depends upon the metallurgical structure through the mean ferrite path, the fracture stress appears to be connected with structure by means of the size of certain structural elements. It is well known that the ductility of steel depends upon the amount, distribution, and size of inclusions and hence, depends upon the amount and direction of hot working. It is also known that in the case of tempered martensitic steels, that the smaller the carbide particles the higher is the fracture stress. The suggestion(161, 165) that the tensile stress required for fracture depends upon the average size of inhomogeneities projected in a plane perpendicular to the applied tensile stress, explains, qualitatively at least, many of the observations and offers a picture that is helpful in interpreting puzzling data. In the case of steels, both the inclusions and the carbide particles are operative in affecting the fracture stress. The size of the inhomogeneities considered in a plane perpendicular to the applied tensile stress changes as deformation proceeds, giving rise to a dependence of fracture stress upon deformation. In the rolling operation, for example, in which the inclusions are aligned in the direction of rolling, the fracture-stress curve for loads applied parallel to the rolling direction is raised, while that for loads applied transversely is lowered. If a single sample of steel is considered (the inclusions constant), the fracture stress should depend on the size of the carbide particles: in the case of tempered martensite on the size of the carbide spheroids, and in the case of pearlite on the largest dimension of the carbide plates*. During deformation of tempered martensitic steels the projected size of carbide particles (in a plane perpendicular to the applied load) does not change appreciably as the deformation proceeds inasmuch as the particles are spheroidal. In the case of pearlitic steels, orientation of the cementite plates occurs during deformation and alters the fracture stress markedly as deformation proceeds. (Figure 31.) For deformation in simple tension, this orientation of the cementite plates results in an increase of fracture stress in the direction of the applied load and a decrease in directions perpendicular to the load. This anisotropy of fracture stress is important. As will be discussed later, many experiments illustrating it have been performed.

The important difference between the fracture-stress curve of tempered martensitic and of pearlitic steels is illustrated best, but indirectly, by the results of notched-bar impact tests, which are discussed in the forthcoming chapter. The fracture-stress curves cannot be measured directly except in certain special circumstances. However, for pearlitic steels, which, as we shall see, break brittlely in simple tension at low testing temperatures, the fracture-stress curves may be determined indirectly(166). Thus, if specimens of a pearlitic steel are first deformed various amounts at room temperature and then cooled to a temperature so low that they break brittlely, the fracture stress (at the low temperature) may be determined as a function of the deformation at room temperature. If it is assumed that the slope of the fracture-stress curve is

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*Bainite will be discussed later.

independent of fracturing temperature, the fracture-stress curve at room temperature can then be constructed*. It is important to point out that the separation of the flow and fracture stresses at a strain of about .01 has been found in this manner to be only about 10,000 p.s.i. for several pearlitic steels(166).

The fracture-stress curves of tempered martensitic steels cannot be determined even by this indirect method, for if these steels have sufficient ductility to permit deformation at room temperature, they do not break brittlely even at very low testing temperatures. However, as will be shown, the mere fact that tempered martensitic steels do not break brittlely under these severe conditions is sufficient evidence that their fracture-stress curves are much less dependent on strain than are the curves for pearlitic steels**. Schematic fracture-stress curves for steels of pearlitic and of tempered martensitic structures are illustrated in Figure 32.

Because the fracture stress of pearlitic (and bainitic) steels is dependent upon strain, the effect on their fracture-stress curves of changes of structure are particularly difficult to determine by the usual tensile tests. If the strain to fracture (reduction of area) changes due to changes in metallurgical structure or in test conditions, it is difficult to determine whether the shape or the level of the fracture-stress curve has been changed or whether the change in measured fracture stress is due simply to the change in flow stress. However, certain general conclusions may be drawn. For example, as the strength of (eutectoidal) pearlitic steels is increased by decreasing the carbide plate separation, the observed fracture stress in simple tension increases faster than does the stress required for flow(164). The only explanation of this is that the fracture-stress curve rises as the structure changes from coarse to fine pearlite. The prior austenite grain size appears to have a definite effect on the properties of pearlitic steels(170). The finer the grain size the higher is the fracture-stress curve (and possibly the less is its slope). Perhaps this increase in fracture stress may be accounted for by the decrease in maximum dimension (length) of the cementite plates with decreasing grain size.

If it assumed that the fracture-stress curve of tempered martensitic steels is nearly independent of strain, the observed fracture stresses in the tension test of these steels at several strength levels have specific meaning, even though the strain before fracture changes. In Figure 33 the

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*Other methods of determining the fracture stress as a function of deformation based upon the use of notched bars have been used (167, 168). These methods have been subjected to criticism(169).

**At strains larger than the fracture strain the fracture curves for pearlitic and for tempered martensitic steels having similar flow curves must approach each other.

fracture strength of a steel having a tempered martensitic structure is plotted as a function of the yield strength. These data indicate that the strength σ_y is related to fracture strength σ_f in the following manner (163):

$$\sigma_y = K (\sigma_f)^2 \qquad (3\text{-}6)$$

The value of K depends upon the quality (inclusions) of the steel and the carbon content. Meagre tensile data (borne out by more complete notched-bar results) indicate that the fracture stress for a given strength of the steel increases as the carbon content decreases. Since for a given flow strength (or mean ferrite path) the size of the carbides are larger, the higher the carbon content; this relation is consistent with the concept of fracture expressed above. At any rate, the strain to fracture in simple tension increases as the carbon content decreases. This increase in fracture strain may be accounted for by a decrease in the slope of the flow stress curve or by an increase in fracture stress as the carbon content decreases.

The relation given in Equation (3-6) applies only to tempered martensitic steels. There is reason to believe (Chapters IV and VIII) that the fracture-stress curve of untempered martensite decreases much more rapidly with the initial precipitation of carbide than does the flow stress curve. The first introduction of the carbide particles may cause a marked decrease in fracture stress. During subsequent tempering the relation between flow and fracture stresses appear to be given by Equation (3-6).

Another metallurgical variable that may have a marked effect in lowering the fracture-stress curve of steel is hydrogen. This hydrogen is dissolved in the metal during melting or treatment. The solubility of the hydrogen in ferrite is low and it presumably precipitates as a gas, inducing interval ruptures*. These ruptures may act in the same way as inclusions in lowering the fracture stress and may reduce the deformation before fracture.

The fracture properties of steel depend to a great extent on one other variation of structure that can occur during normal heat treatment. The precipitation that results in temper brittleness does not markedly affect the flow-stress curves but alters the fracture-stress curves profoundly. This precipitation can be made to occur in steels having any metallurgical structure. Therefore, the effect may be imposed on any type of fracture-stress curve. The effect of the precipitate on a steel having a tempered martensitic structure is illustrated schematically in Figure 34. The precipitate lowers the fracture stress at small strains and so, as will be seen, induces brittle failure in notched specimens even at relatively high testing temperatures. In extreme cases the fracture-stress curve is lowered sufficiently to cause a reduction in the strain to fracture in the tensile test.

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*The precipitation of hydrogen and its embrittlement of steel are reviewed in several publications(171).

Effects of External Variables

Complex Stresses

In many engineering applications the stress distribution is complex; metal members of engineering structures are not often subjected to simple tensile stresses. Even if a metal part is subjected to a simple tensile load, stress concentration arising from sharp corners and cracks render the stress distribution quite complex. Often, therefore, it is necessary to calculate the flow or fracture stress of a metal under stress combinations more complex than that of simple tension. This problem is not one finding how the flow or fracture stress changes with deformation but only of finding the conditions that govern flow and fracture under a specific stress pattern for a fixed amount of deformation.

Yielding and Flow*

Many hypotheses have been advanced as to the conditions under which plastic flow occurs under complex stress patterns. The concept that yielding occurs when the so-called "energy of distortion" reaches a critical value is probably the most nearly in agreement with the observed data(173, 174). If a metal is loaded under any complex system of stresses it will flow plastically when:

$$(\sigma_x - \sigma_y)^2 + (\sigma_y - \sigma_z)^2 + (\sigma_z - \sigma_x)^2 = K. \qquad (3\text{-}7)$$

where σ_x, σ_y, σ_z, are the three principal stresses and K is a constant. All then that is necessary to know is the value of the constant K, which may be determined by measuring the stress required for plastic flow under a known combination of stresses. For example, suppose that the stress required to produce some specified small amount of plastic deformation of a steel in simple tension is known and it is desired to find the tensile stress required for flow when the metal is completely constrained from plastic deformation in one lateral direction. From symmetry considerations it can be seen that complete lateral constraint is obtained when the applied transverse stress is exactly equal to one half the longitudinal stress**. In the case of flow in simple tension:

$$K = 2(\sigma'_x)^2 \qquad (3\text{-}8)$$

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*Discussions of the effects of complex stresses on yielding and flow are presented in several treatises(161, 172).

**Under the conditions of constant volume a metal contracts 1/2 as much laterally as it extends longitudinally. A transverse stress of 1/2 the longitudinal will prevent the lateral contraction.

where the superscript ' refers to initial flow. In the case of a single lateral constraint:

$$K = 2(\sigma'_{xc})^2 + 2(\sigma'_{yc})^2 - 2(\sigma'_{xc})(\sigma'_{yc}) \qquad (3\text{-}9)$$

And since:

$$\sigma_{yc} = 1/2\ \sigma_{xc}, \qquad (3\text{-}10)$$

$$K = 3/2(\sigma'_{xc})^2 \qquad (3\text{-}11)$$

where the subscript c refers to the constraint conditions imposed.

Flow under these conditions will occur when:

$$2(\sigma'_{x})^2 = 3/2\ (\sigma'_{xc})^2 \qquad (3\text{-}12)$$

or:

$$(\sigma'_{xc}) = 1.15\ (\sigma'_{x}) \qquad (3\text{-}13)$$

Thus, the tensile stress required for initial flow when a specimen is constrained from plastic deformation in one lateral direction is fifteen per cent higher than the stress required for plastic flow in simple tension.

Other more complex cases may be computed from Equation (3-7), which is very nearly in agreement with the observations for almost all metals. During the inhomogeneous yielding of some steels, it is questionable whether or not the expression given in Equation (3-7) precisely fits the facts(175). It is certainly sufficiently accurate for most engineering applications.

Fracture

The criterion for fracture under complex stresses is very difficult to formulate, for the tensile stress required for fracture is a function of the amount and type of deformation preceding fracture. Therefore, in any series of experiments in which the strain before fracture changes as the type of loading changes, the fracture stress will vary but not necessarily because of the change in type of loading. Combined stress tests have been performed (148,176) on hollow tubes of various types of metals, including single phase alloys such as copper and brass which should show little dependence of the fracture stress on strain. These experiments indicate that for a given metal, fracture occurs when the maximum tensile stress reaches a critical value independent of the magnitude of transverse stresses. Further, results of notched-bar impact tests indicate that the fracture stress-strain curve is not affected by a single transverse tension. In general, the law that fracture occurs at a fixed tensile stress can hold only for a metal deformed a given amount by a given type of deformation, for the fracture stress is dependent upon deformation and is anisotropic.

Complex Strains

The discussion above is concerned only with the stress required for initiating flow or for fracture under any particular combination of stresses. If a metal is deformed under complex loading conditions, it is often necessary to know how the flow and fracture stresses vary with the deformation. In this connection, it is interesting to discover the effects of reversed strains, such as tension followed by compression and of torsion followed by tension, etc. A comparatively large amount of information is available concerning the strain-hardening in the most general types of deformation but fracture studies are extremely difficult and have been sadly neglected.

Yielding and Flow

It has been found that there exists a generalized expression for the flow stress and a generalized expression for the strain* that are uniquely related for the deformation of most metals, at least for moderate strains. The formulation of an unique relation between stress and strain assumes that strain-hardening is isotropic**. Divergences from the unique relation between stress and strain are, therefore, reflections of anisotropy. Generally speaking, however, such anisotropy effects are not important in engineering design and are interesting primarily to students of the behavior of metals. Furthermore, the complex expressions for the generalized stress and strain are in themselves not easy to apply except in special cases in which the strains can be obtained from geometric considerations alone.

It must be remembered that strain should be defined in the Ludwik sense, that is, the length over which the strain increment occurs must be properly chosen. With the strain defined in this manner, the approximate isotropy of strain-hardening permits the simple addition of generalized strain. For example, suppose that a metal is strained in compression to a strain of .2. If now the compression is followed by a tensile strain and the metal is returned to its original shape, the total strain will be .4 and the strain-hardening will be very nearly the same as if the total strain had been a simple compressive or tensile strain of .4.

To illustrate the relations between generalized stress and strain and the range of strain over which they are valid, a comparison can be made between strain-hardening in tension and in torsion. In this particular case the tensile stress T and strain ϵ in the tensile case are related to the generalized stress σ_I and strain ϵ_I by the following equations(161):

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*These relations are presented and discussed in a recent paper(161).

**It is also assumed that the energy of distortion criterion is applicable to plastic flow. As discussed above, this criterion has been shown to be very nearly in accord with the facts.

$$\sigma_I = (\sqrt{2}/3)T$$
$$\varepsilon_I = (3/\sqrt{2})\varepsilon \quad \text{(Tension)} \qquad (3\text{-}14)$$

The shear stress, S, and shear strain, s, in the torsion case are related to the generalized stress and strain by the following equations:

$$\sigma_I = \sqrt{2/3}\ S$$
$$\varepsilon_I = \sqrt{3/2}\ s \quad \text{(Torsion)} \qquad (3\text{-}15)$$

Therefore, the stress-strain curves in torsion and tension should be correlated as follows:

$$(S,\ s) = (T/\sqrt{3},\ \varepsilon\sqrt{3}) \qquad (3\text{-}16)$$

According to this viewpoint, if the tensile stress is divided by and the tensile strain is multiplied by the square root of three, the resulting strain-hardening curve should superimpose on the curve for torsion if the material is isotropic. In Figures 35 and 36 logarithmic curves of this type are plotted for a tempered martensitic steel and for a pearlitic steel. The divergence between the tensile and torsion curves occurs sooner for the pearlitic than for the tempered martensitic steels, possibly indicating that orientation of inhomogeneities at large strains modifies the stress-strain relation for flow as well as for fracture. This orientation effect may give rise to the linearity of the tensile stress-strain curve for large strains. However, the tensile and torsion curves agree up to moderate strains. Thus, tensile stress-strain curves may be utilized to determine for small strains, the strain-hardening characteristics when the strain is other than a simple tensile strain. Further, because of the approximate isotropy of strain-hardening, the strengthening arising from any combination of strains may be computed if the strains are properly defined and a single flow-stress curve is known.

In the case of reversed strains such as tension followed by compression, it is found that the stress required for the initial plastic flow following the reversal (initial flow in compression) is always lower than would be predicted by the flow-stress curve. After an initial tensile deformation, the application of a compression load causes plastic flow to occur almost immediately upon the application of the load, and the flow-stress curve is curved almost from zero load. After a comparatively small amount of plastic deformation, the flow curve becomes the same as that for compressive deformation not following tension. This behavior has been studied by many investigators and is referred to as the Bauschinger effect (178). The effect arises from variations on a microscopic scale of the stress required for plastic flow. During tensile deformation plastic flow occurs in some regions of the metal at lower stress than others. Removal of the load leaves in compression those regions that flowed at the lower stress. If now a compression follows tension, the regions of the metal that were left in compression will yield at a low value of the average compression stress. A low temperature heat-treatment or a little plastic flow will remove the stress peaks and the Bauschinger effect.

The Bauschinger effect is connected with the anelastic effects (103) mentioned earlier. Another example of anelasticity is the elastic after-effect(179). After a metal has been deformed and the load is removed, the specimen retracts immediately to some extent. The specimen continues to retract slowly until the total anelastic deformation is relieved. This gradual contraction after removing a tensile load is known as the "elastic after-effect". It is connected with the relief of strains on a microscopic scale by "relaxation" effects. If these microstresses are not given time to relax, then an application of a reversed load gives rise to a Bauschinger effect. At room temperature the microstresses require very long times for removal. A low temperature heat-treatment will accelerate the process and cause the disappearance of at least a part of the Bauschinger effect.

Fracture

The available evidence indicates that fracture occurs in a given direction for any given metal when the tensile stress in that direction reaches a critical value. If fracture does not occur but plastic flow takes place, it is known that the stress required for fracture changes and changes differently in every direction. Thus, the effect of deformation on the tensile stress required for fracture is anisotropic. No general laws have been proposed relating the variation of fracture stress with deformation. In fact, since the dependence of fracture stress on deformation varies with metallurgical structure, it would appear difficult to formulate general laws. Specific examples will probably serve as the best illustrations of this dependence of fracture stress on deformation.

Experiments have been performed(180) that indicate that if a pearlitic steel is first twisted and then deformed in tension, fracture occurs in tension not on a plane perpendicular to the applied stress but on a helical plane that was under compression during the torsion. The concept of the orientation of inhomogeneities is quite consistent with this behavior(161). As the deformation in torsion proceeds, the inclusions and the cementite plates of the pearlite tend to become oriented toward the helical compression surface and the tensile stress necessary for fracture perpendicular to this plane is reduced. Thus, a specimen deformed in torsion may, if then deformed in simple tension, break on the helical surface that was under compression in the torsion test. If after twisting in torsion, the specimen is untwisted, the inhomogeneities tend to revert to their original configurations. On subsequent tension(180) the fracture reverts to a typical cup-cone and the strain to tensile fracture increases. The untwisting does further strain-harden the steel. It appears, therefore, that the fracture stress does not depend upon strain in the way that flow stress does, but rather depends upon the orientation of inhomogeneities during deformation.

If a metal is deformed in compression the metal strain-hardens as indicated in Figures 22 or 23. Unless bulging of the specimen occurs, however, there are no tensile stresses present and fracture will not occur. During the compression of a metal whose fracture stress is very dependent upon tensile strain, the tensile stress required for fracture on a plane

perpendicular to the axis of the compression specimen may decrease as indicated in Figure 37. For compressive strains greater than $\mathcal{E}_c$, the subsequent application of a tensile load will cause failure without deformation. The presence of the Bauschinger effect may cause the minimum compressive strain for subsequent failure without deformation in tension to be slightly greater than $\mathcal{E}_c$. At any rate, a pearlitic steel deformed in compression should subsequently break in tension at a much smaller strain than that encountered in a simple tension test of the same metal ($\mathcal{E}_t$). This prediction is borne out in experiments by recent investigators(181).

If a pearlitic or temper brittle specimen is deformed in tension, the tensile stress required to fracture the specimen on a plane perpendicular to the axis increases while the stress required to fracture the metal in a plane parallel to the axis of the specimen decreases. This decrease in fracture stress sometimes leads to longitudinal fractures in the tensile test (Chapter IV).

Hot-rolling of steel plates causes an orientation of the inclusions in the direction of rolling and in a plane parallel to the plate. It is observed that, even after subsequent heat-treatment, the fracture stress of specimens taken from the plate parallel to the direction of rolling is higher than for transverse specimens, while specimens taken perpendicular to the plate have by far the lowest fracture stress.

The variation of the effective size of inhomogeneities in a plane perpendicular to the applied tensile stress explains, qualitatively at least, all the known variations of fracture stress with strain.

Temperature

The effect of temperature on the stress-strain curves (tensile or generalized) is two-fold. Not only does an increase in temperature decrease the height and change the shape of the flow and fracture curves, but changes in temperature may also cause changes in metal structure to occur simultaneously with the deformation. In steels, for example, a precipitation probably* takes place in strained material at temperatures between about 70°F. (20°C.) and 400°F. (200°C.) that alters the structure of the steel. Further, if the temperature of steel is raised sufficiently, the normal structural changes such as tempering, austenite formation, etc., will occur. At high temperatures, recovery and recrystallization also occur concurrently with the strain-hardening. Lowering the temperature below about 70°F. (20°C.) however, results only in an increase in strength with little possibility of the occurrence of transformations that affect the structure of the steel. Since the effects of high temperature on the stress-strain curves are so complex and the experimental procedures so difficult, there is very little direct information on them.

*It is not definite that the strain-aging to be described in the following is due to precipitation though this mechanism is extremely likely.

Yielding

As the temperature is lowered the initial heterogeneous type of deformation becomes more pronounced. Since the lower yield stress rises with decreasing temperature faster than does the flow curve, the lower yield strain increases. Furthermore, steels of moderate yield strength that do not exhibit the Piobert effect when tested at room temperature may begin to yield inhomogeneously as the temperature is lowered(182).

Flow

Only a very few experiments from which complete flow-stress curves may be obtained have been performed at low temperatures. Data on one steel for both tempered martensitic and for pearlitic structures, at two strength levels, have, however, been obtained(182). Some of these data are plotted in Figure 38. In the case of the pearlitic steel the stress σ_ε required to produce a given plastic strain ε is related to temperature in the following fashion over the whole temperature range of the tests:

$$\sigma_\varepsilon = Ke^{q/RT} \qquad (3\text{-}17)^*$$

where K and R are constants, q a constant that depends upon the value of σ_ε at a given temperature, and T the absolute temperature. Since, as will be discussed subsequently, the tensile strength (Chapter IV) is a fairly good measure of the height of the flow-stress curve, the variation of the tensile strength with decreasing temperature is illustrated in Figure 39 for a number of pearlitic steels. These data indicate that:

$$q = \frac{C}{S_o^{\,3/4}} \qquad (3\text{-}18)$$

where C is a constant and S_o the tensile strength at 70°F. (20°C.). From this expression the variation of the height of the flow-stress curve (in this case the tensile strength) with temperature may be estimated. A chart showing the variation of the tensile strength with temperature is presented as Figure 40.

For tempered martensitic steel, equation (3-17) holds but the value of q appears(182) to change abruptly at about -75°F. (-60°C.). Recently, data illustrating the effect of temperature on the tensile strengths of tempered martensitic steels have been obtained(184) that also show a discontinuity in the vicinity of -75°F. (-60°C.).

It appears that for all metals there are two distinct ranges of temperature in which the dependence of the flow-stress curves on temperature is widely different. There are very few data on the variation of the stress at small strains for steels over the whole range of temperatures

*The lower yield strength is related to temperature in the same manner but the value of q is larger.

from very low up to 1300°F. (700°C.). Figure 41, which shows the dependence of the torsional strength* of several pearlitic steels, is typical, however**. In both ranges of temperature the stress required to produce a given plastic strain varies with temperature according to Equation (3-17) but the value of σ is widely different above and below about 650°F. (350°C.). The presence of these two regions of temperature produces a curve of the type illustrated in Figure 42 when the flow stress at a small strain is plotted directly against temperature***. In the case of steels the stress required to produce a small plastic strain varies with temperature in this way, but as discussed below, the stress at larger strains decreases, increases, then decreases as the temperature is raised above room temperature.

As the temperature is decreased below 70°F. (20°C.) the slope of the entire stress-strain curve increases more rapidly than if the strength were increased by heat-treatment. Both the strain-hardening exponent**** and the strain-hardening modulus increase, in steels of all structures, whereas when the strength is increased by heat-treatment, the strain-hardening exponent decreases.

If the temperature is raised above room temperature the slope of the stress-strain curve for steels at first decreases, but at temperatures of about 400°F. (200°C.), it begins to increase (at strain rates of the order of 10^{-4} sec.$^{-1}$). The stress-strain curves obtained at higher temperatures actually cross those for lower temperatures(187). As has been pointed out earlier, the stress required to produce a small plastic strain decreases continuously with increasing temperature. This effect of deformation and temperature is illustrated schematically in Figure 43. Concurrent with this increase in strain-hardening the stress-strain curves become serrated and apparently consists of consecutive small inhomogeneous deformations. This phenomenon is connected with the Piobert effect. It is known,

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*The maximum torque reached in the torsion test (a measure of stress required for flow at moderately large stresses) is generally called the torsional strength.

**The slight rise in the curve above 400°F. (200°C.) is due to an increase in strain-hardening possibly due to precipitation, discussed in the following pages.

***The slight inflection in the curve has been attributed to aging, in the case of non-ferrous metals(186).

****This increase in strain-hardening modulus has been observed for many steels and is responsible for the increase in per cent elongation to fracture often observed as the temperature is lowered (Chapter IV).

for example, that if a steel having a drop in load at yielding is deformed by an amount greater than the lower yield point, aging at room temperature will cause the recurrence of the inhomogeneous yielding(154). This serration can be explained if precipitation of carbide or nitride is taken as the mechanism of strain aging. At elevated temperatures precipitation can occur more quickly, causing a reappearance of the lower yield phenomenon and an increase in flow strength. The continuation of deformation and precipitation results in a serrated stress-strain curve(187, 190). Above temperatures of about 700°F. (350°C.) the strain-hardening decreases continuously with increasing temperature.

Fracture

Because of the variation of the fracture stress with strain, little can be said about the effect of temperature upon it. However, by finding both the variation of the ordinarily measured tensile properties with temperature (below room temperature) and the effect of deformation, recent investigators(166, 182) were able to determine approximately the effect of temperature on the height of the fracture-stress curve. The effects of temperature upon the flow strength, upon the fracture strength (at different fracture strains), and upon the strain to fracture (reduction of area), in simple tension, are illustrated in Figure 44 for a pearlitic steel. In Figure 45 is illustrated* the variation of the fracture stress with strain, obtained by deforming specimens various amounts at room temperature and breaking them at -310°F. (-190°C.). From the variation of the reduction of area (fracture strain) of Figure 44 and the effect of this reduction of area in increasing the fracture stress (Figure 45), the stress required for fracture at any temperature may be determined**. In Figure 44 the variation of the fracture stress at a strain of .01 with temperature is also plotted. It is evident that the fracture stress varies with temperature according to Equation (3-17) but the value of α is smaller than for the flow stress at the same strain. Thus, the fracture-stress curve rises with decreasing temperature but less rapidly than does the flow-stress curve.

In the case of tempered martensitic steels for which the dependence of the fracture stress curve on strain is assumed to be small, the approximate variation of the height of the curve may be obtained directly from the tensile test data. Such curves are presented in Figure 46. In this case we see that the fracture curve rises less rapidly than the flow curve and there is a discontinuity at about -75°F. (-60°C.). The cause of this abrupt change may be probably associated with a change in type of fracture(183).

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*These experiments were also performed by another investigator(192).

**In calculating the fracture stress at a constant strain it is assumed that the slope of the fracture stress vs. reduction of area curve (Figure 45) is independent of fracturing temperatures. The independence of the fracture-stress curve fracture temperature (below 70°F.) has not been demonstrated(193).

Rate of Strain

Extremely large changes in strain rate are necessary to induce even small changes in the flow stress curves. Relatively few researches have been performed covering a sufficiently wide range of strain rate to permit definite conclusions as to the exact quantitative relation between the height or shape of the flow-stress curve and the rate of strain. A limit to the maximum strain rate that can be employed for such measurements is set by elastic vibrations and by inertial effects induced by impacting specimens at high velocities(194). Tests performed at high rate of strain must, therefore, be analyzed with caution.

Yielding

Increases in strain rate raise the lower yield strength faster than the flow curve. The lower yield elongation increases as the strain rate is increased, as it does when the temperature is decreased. It has been found that the effect of strain rate on the lower yield strength can be expressed by the following relation:

$$\sigma = K(\mathcal{E})^n \qquad (3\text{-}19)$$

where σ is the stress, $\mathcal{E}$, the rate of strain, K a constant and n another constant that depends upon the strength of the steel (at a fixed strain rate) (183). Other relations between strain rate and strength that appear to fit the observed data almost as well, have been suggested(195). Since, however, the effect of strain rate is so small (the power, n, is small) it is difficult to determine which relation is correct.

Data illustrating the above relation between the lower yield strength and strain rate are plotted as Figures 47 and 48. The higher the lower yield strength initially, the less is the effect of strain rate (the lower is the strain rate exponent).

Flow

The flow stress (on the rising part of the flow-stress curve) appears to increase with strain rate in the same manner as does the lower yield strength (Equation 3-19). The strain rate exponent is less for the flow stress than for the lower yield strength (for steel of the same strength level). At large strains the stress required for flow appears to rise faster with strain rate than does the stress at small strains(182). Thus, the effects of increasing strain rate are similar to the effects of decreasing temperature. As the strain rate is decreased (at say 70°F.) the slope of the flow-stress curve first decreases, then rises again(187) and then further decreases. This increase in the slope of the flow-stress curve occurs in exactly the same way as when the temperature is raised (strain aging). The stress-strain curves are also serrated(187).

As the strain rate is increased, the deformation tends to become adiabatic; the heat generated by the plastic deformation does not have time to dissipate. At large strains the adiabacity results in an increase in

temperature of the specimen, which generally causes a decrease in the height of the flow-stress curve. Adiabacity of deformation frequently has important consequences(182, 188).

Fracture

The effects of strain rate on the fracture-stress curves have not been determined directly but, as will be seen in the following, it appears that increasing the rate of strain is qualitatively equivalent to decreasing the temperature.

Relation Between Effects of Strain Rate and of Temperature

One example of the qualitative equivalence between the effects of strain rate and temperature is illustrated by their effects on the strain-aging phenomenon. As the temperature is increased the flow stress at large strains decreases, then increases due to strain-aging, and then decreases again. The variation of the flow stress curve with strain rate is analogous. This relation between the effects of strain rate and temperature is illustrated in Figure 49. The higher the strain rate, the higher must the temperature be to induce appreciable precipitation and the higher in temperature is the "hump" in the tensile strength* vs. temperature curve.

A quantitative(161, 182) as well as a qualitative relation(199) between the effects of temperature and strain-rate has been suggested. The quantitative relation is of the form:

$$\sigma_\varepsilon = k(\dot{\varepsilon}\, e^{Q/RT})^n \qquad (3\text{-}20)$$

where σ_ε is the stress at a given strain, $\dot{\varepsilon}$ the strain rate, R and k constants and Q and n constants that depend on the steel. This relation was proposed only for that range of temperature and strain rate in which no changes of metallurgical structure occur. It is to be noted that if either the temperature or the strain rate is constant this equation is equivalent to Equation (3-17) or (3-19). The significance of this relation is that the stress required for plastic flow depends only upon the parameter:

$$P = \dot{\varepsilon}\, e^{Q/RT} \qquad (3\text{-}21)$$

and not upon the strain rate and temperature independently. Thus, a given stress may be obtained at a high strain rate and a high temperature or at a low strain rate and a low temperature. This relation for flow has been tested(182) over a wide range of temperature and a narrow range of strain rate**. In Figure 50 the flow stress required to produce a small strain

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*The tensile strength is a measure of the height of the stress-strain curve at moderate strains and reflect the effects of strain-aging.

**It should be pointed out that this relation holds only for isothermal deformation or for small strains.

is plotted as a function of the parameter P. The constant Q was found to have the value of about 10,000 cal./gm. mol. for the several steels investigated. At least to a first approximation, Equation (3-20) appears to describe the relations between flow stress, strain rate, and temperature. Thus, the shape of the flow-stress curve is expected to change with increasing strain rate in the same way that it does with decreasing temperature.

It has already been pointed out that the temperature of maximum strain-aging depends upon the rate of strain. The relation between the temperature and the strain-rate for maximum strain-aging is given by Equation (3-21), as illustrated in Figure 51.

Figure 52 illustrates the equivalence between strain-rate and temperature. This chart may find practical use if it be assumed that the value of Q lies between 5000 and 15,000 cal./gm. mol., within which limits the value has been found to lie for the steels which have been tested*. If, say it is known that a change in temperature from 70°F. (20°C.) to -75°F. (-60°C.) induces a change of flow strength of 10,000 p.s.i., a steel having a Q of 10,000 cal./gm. mol. would require an increase of strain rate of 10^3 to produce the same effect.

At higher temperatures there still appears to be a relation between temperature and rate of strain necessary to produce a given stress. A given stress is obtained with a given value of the parameter P; however, the value of Q depends upon the stress itself(177, 183). The change of the value of Q may be related to changes of structure by the elevated temperature.

Only one set of experiments happen to have been performed from which the relation between temperature and strain-rate for fracture can be established. Certain experimenters(200), by performing impact tests, found the relation between the temperature and strain rate for brittle failure. They found the relation to be exactly that given by Equation (3-21) (Figure 53).

*The value of Q has actually been determined only for a few steels.

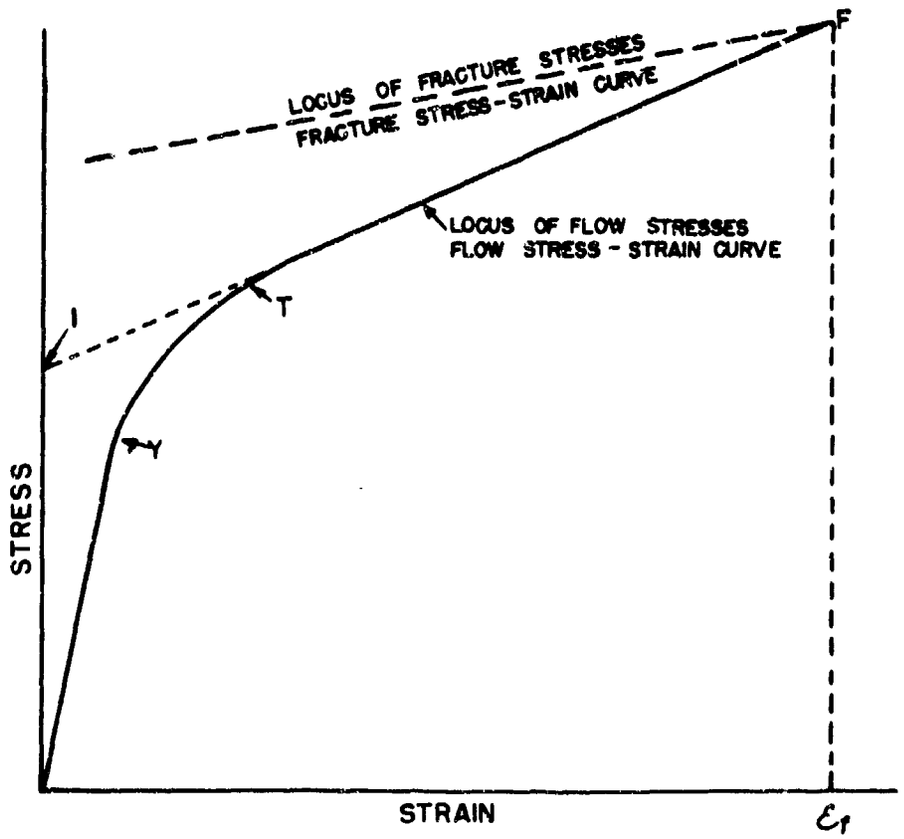


FIGURE 22

FLOW AND FRACTURE STRESS-STRAIN CURVES
(SCHEMATIC)

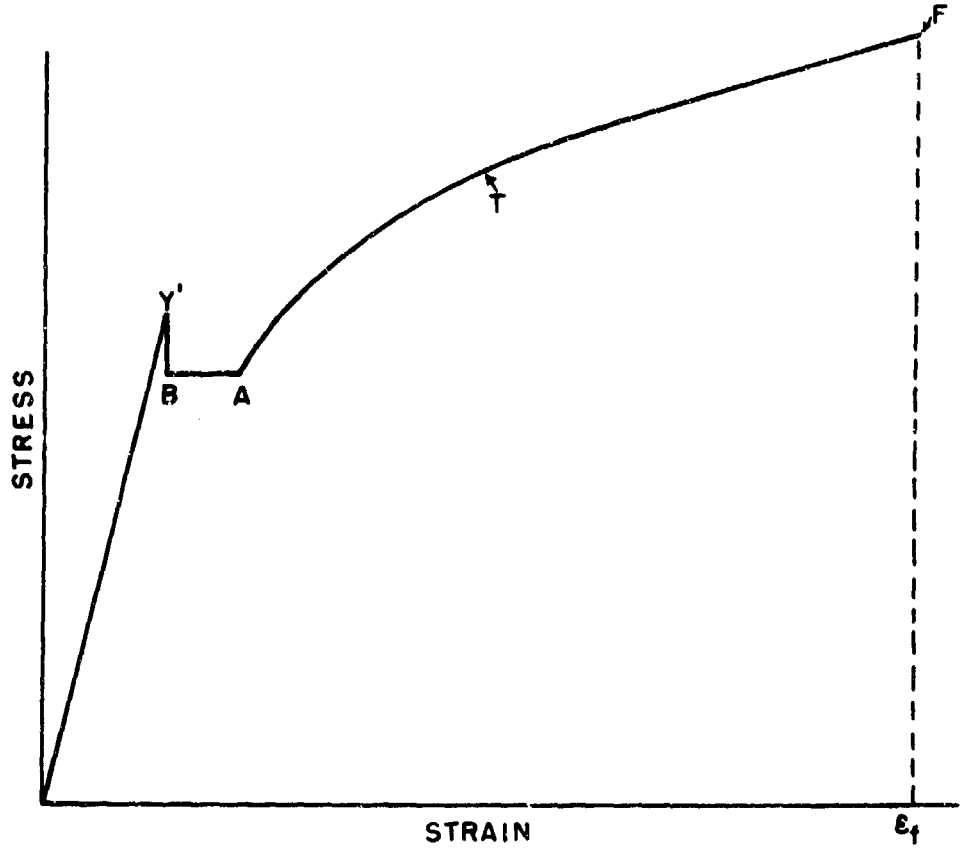


FIGURE 23

FLOW-STRESS CURVE FOR METALS YIELDING INHOMOGENIOUSLY

FIGURE 24

LÜDERS' BANDS IN STEEL

(FROM CARPENTER AND ROBERTSON(151))

WTN.639-7565 1

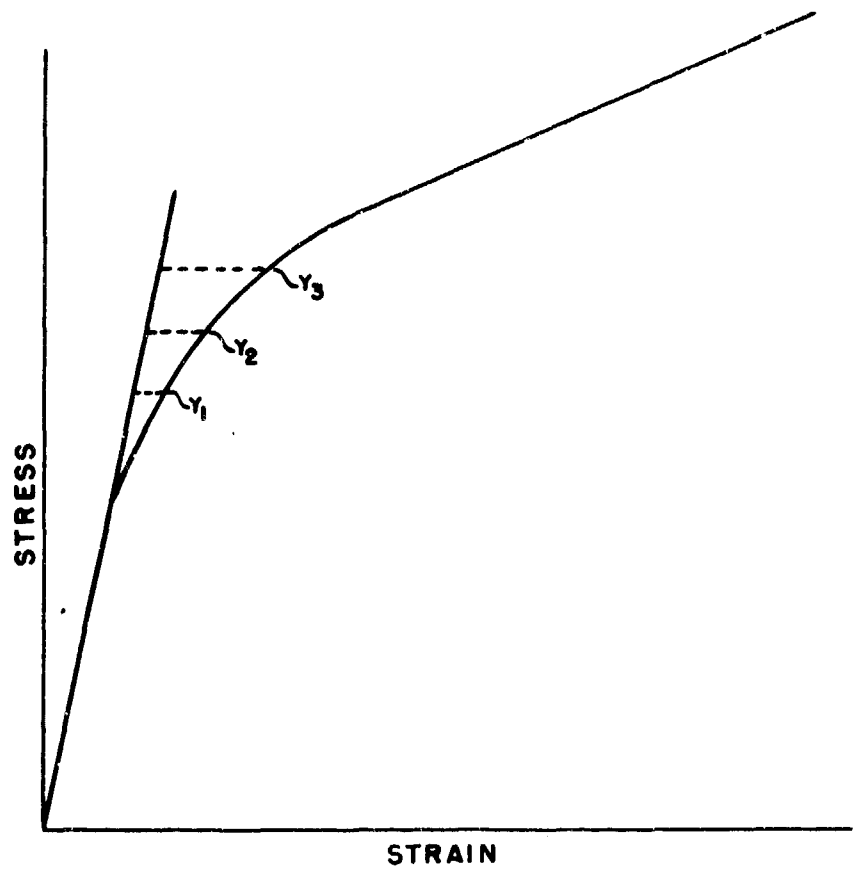


FIGURE 25

EFFECT ON FLOW STRESS
CURVE OF AN INCREASE OF
LOWER YIELD STRENGTH. (SCHEMATIC)

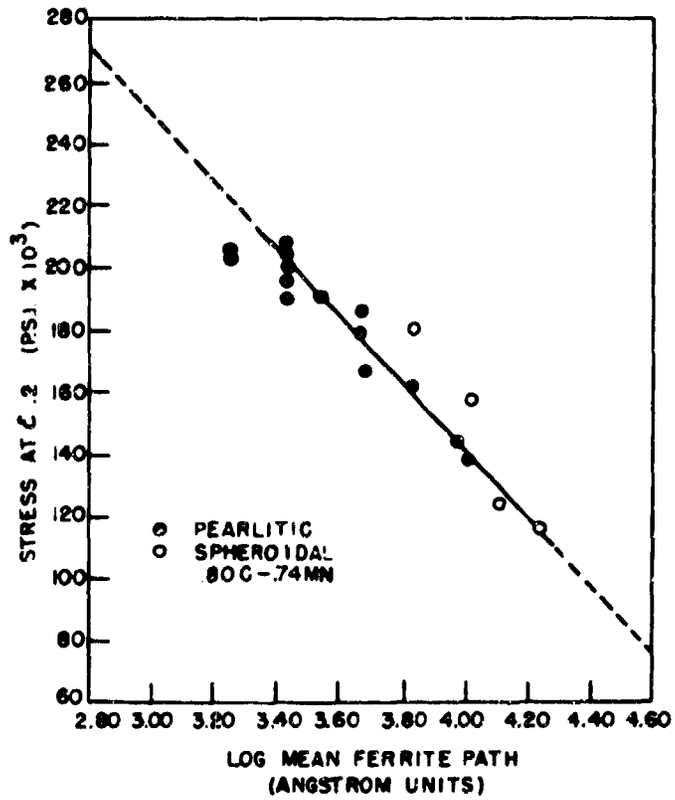


FIGURE 26

RELATION BETWEEN STRENGTH & MEAN FERRITE PATH (AFTER GENSAMER et al[184])

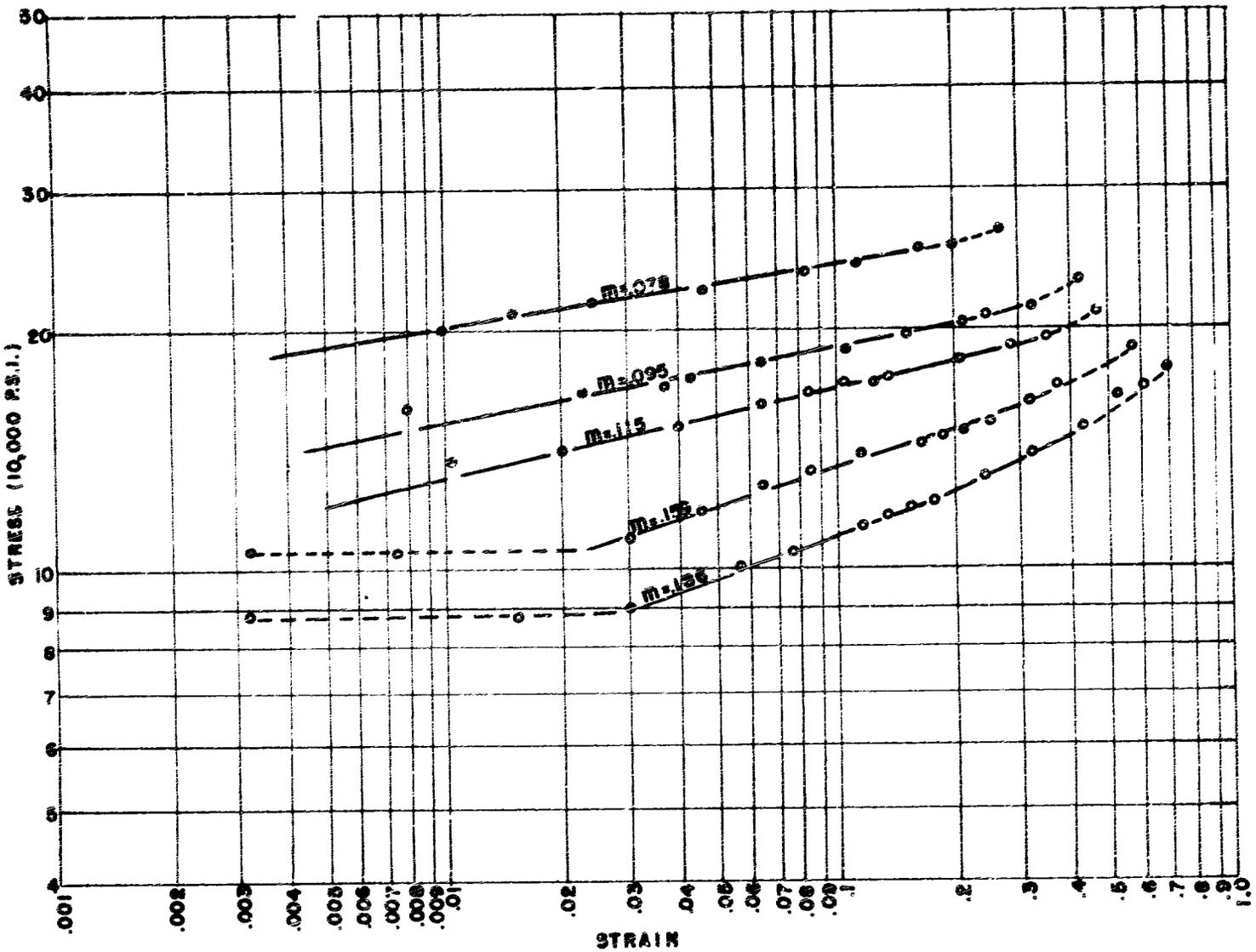


FIGURE 27

LOGARITHMIC FLOW STRESS CURVES OF AN ALLOY STEEL AT VARIOUS STRENGTH LEVELS. (AFTER HOLLOMON[153])

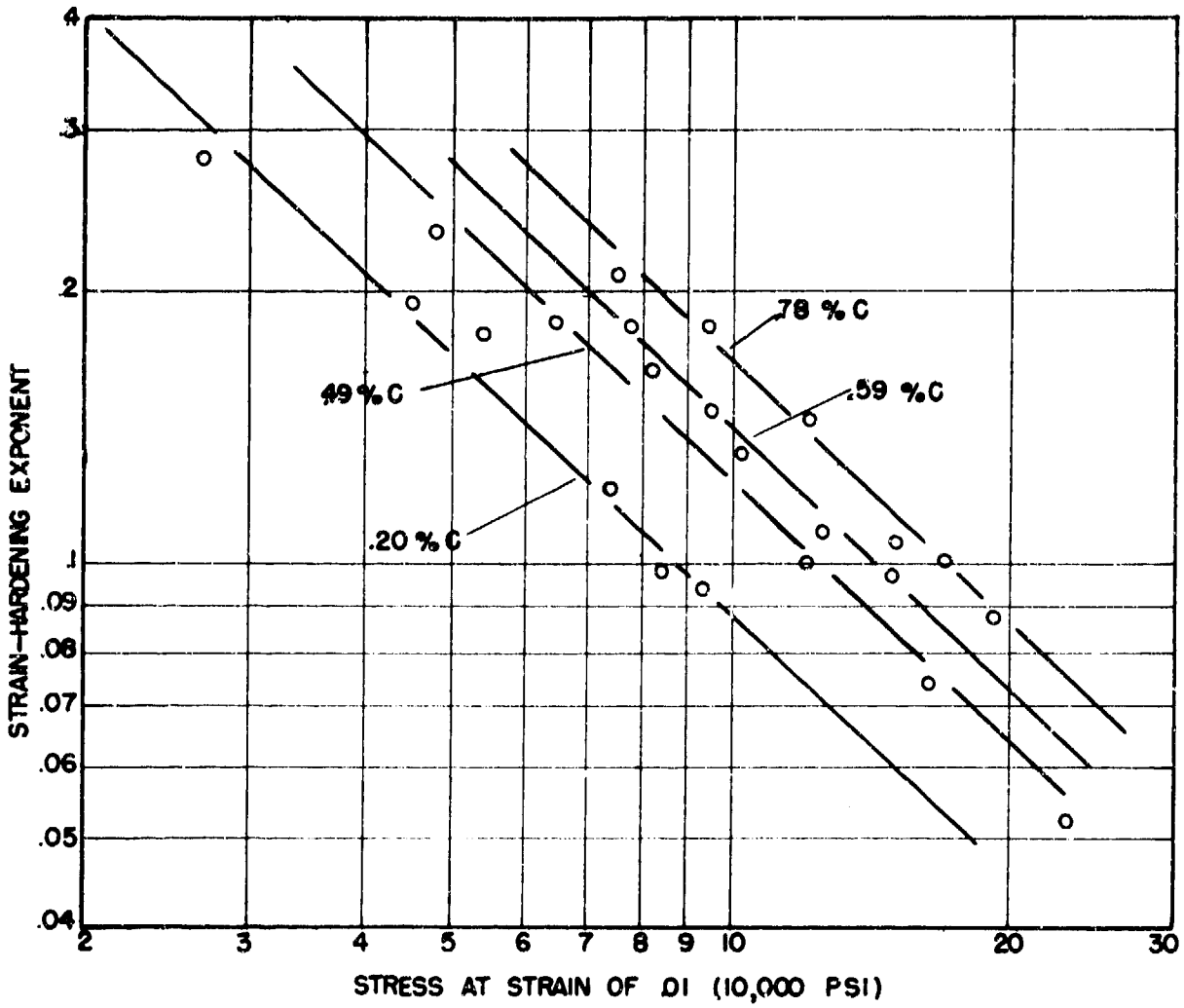


FIGURE 28

EFFECT OF CARBON CONTENT AND STRENGTH LEVEL ON STRAIN-HARDENING EXPONENT FOR PLAIN CARBON STEELS. (AFTER HOLLOMON)[165]

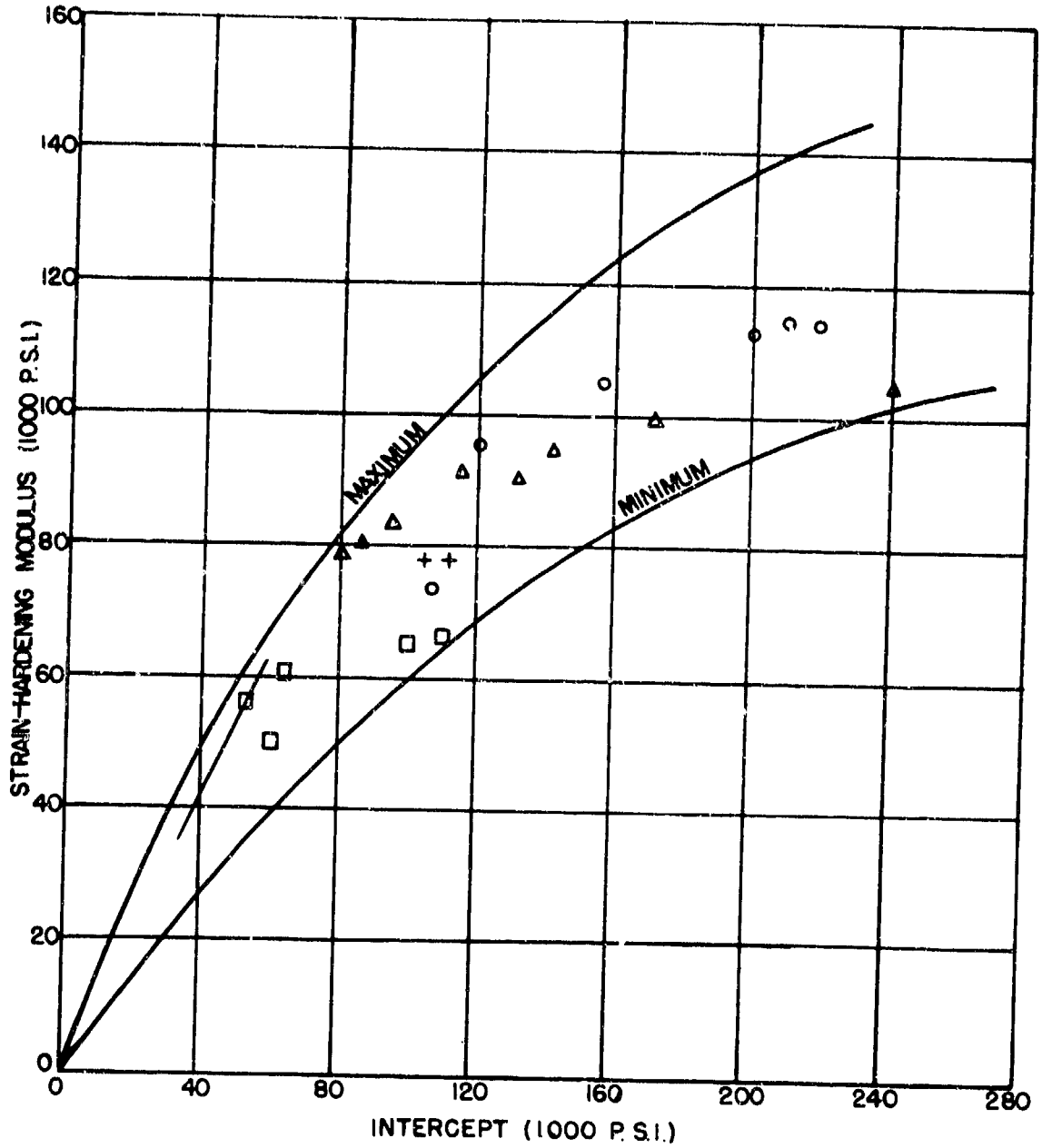


FIGURE 29

RELATION BETWEEN STRAIN-HARDENING MODULUS & STRENGTH INTERCEPT. (AFTER HOLLOMON[163])

(DIFFERENT POINTS REFER TO DIFFERENT STEELS; STRAIGHT LINE REFERS TO ALLOYED FERRITES.)

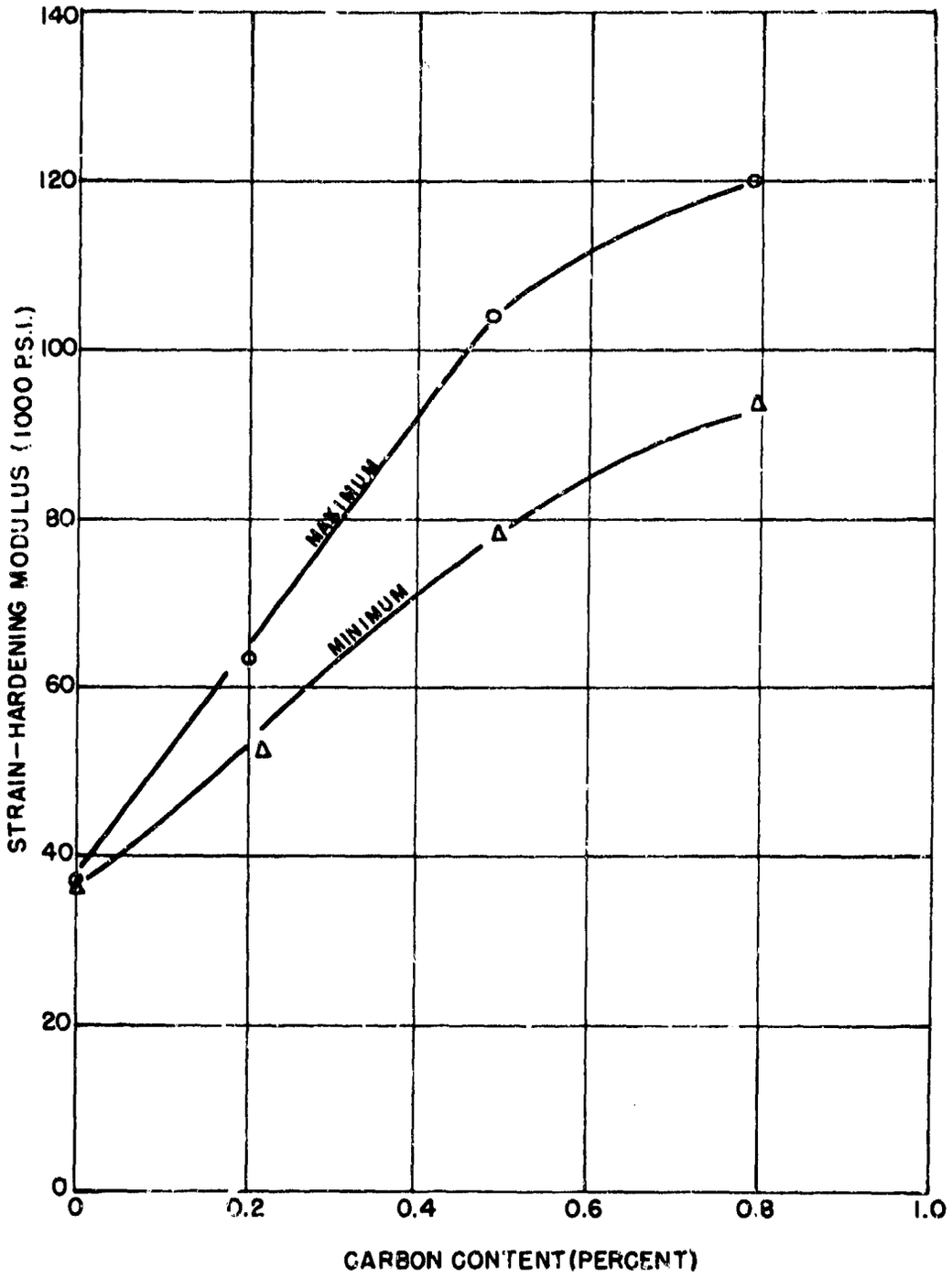


FIGURE 30

RELATION BETWEEN STRAIN-HARDENING MODULUS AND CARBON CONTENT (AFTER HOLLOMON[163])

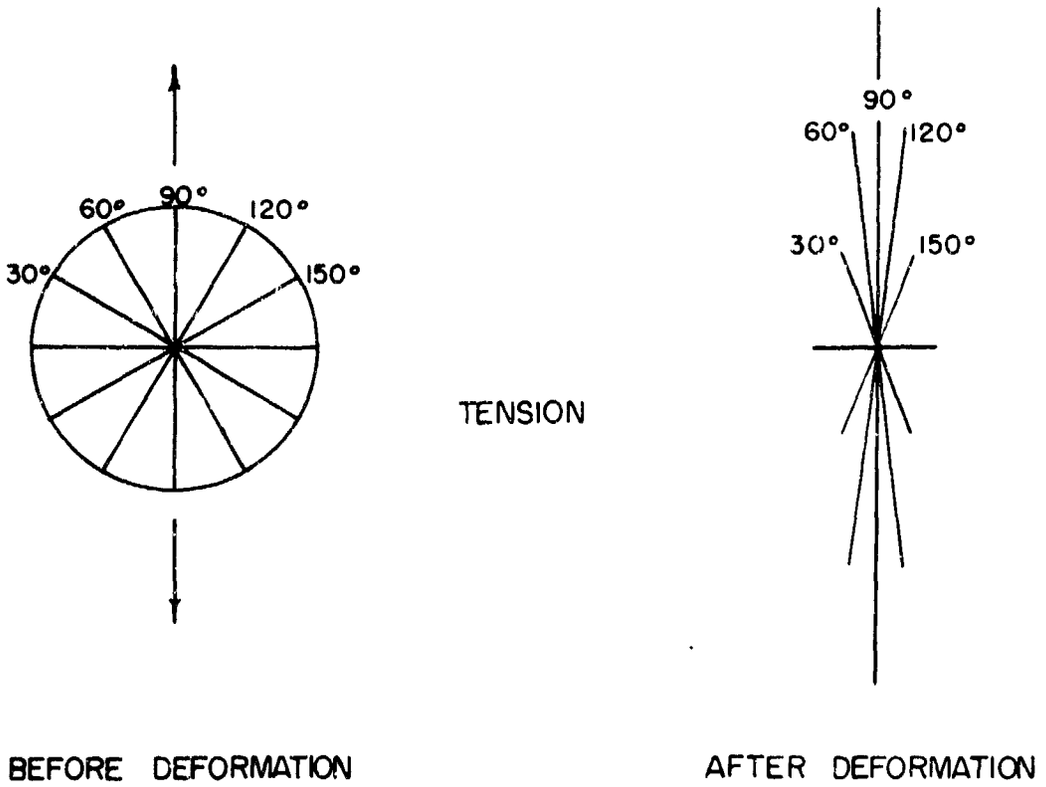
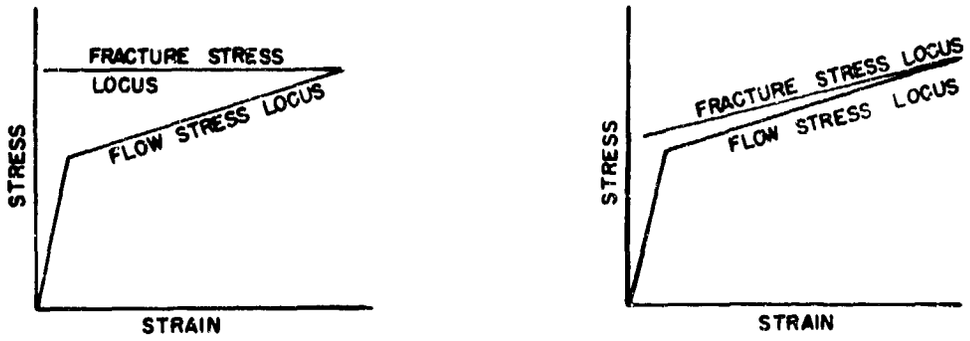


FIGURE 31

ORIENTATION OF CEMENTITE PLATES AND INCLUSIONS DURING DEFORMATION IN TENSION.

(AFTER ZENER & HOLLOMON[161])



TEMPERED MARTENSITE

PEARLITE

FIGURE 32

TENSILE FLOW-AND FRACTURE-STRESS CURVES FOR STEELS HAVING TYPICAL STRUCTURES. (SCHEMATIC)

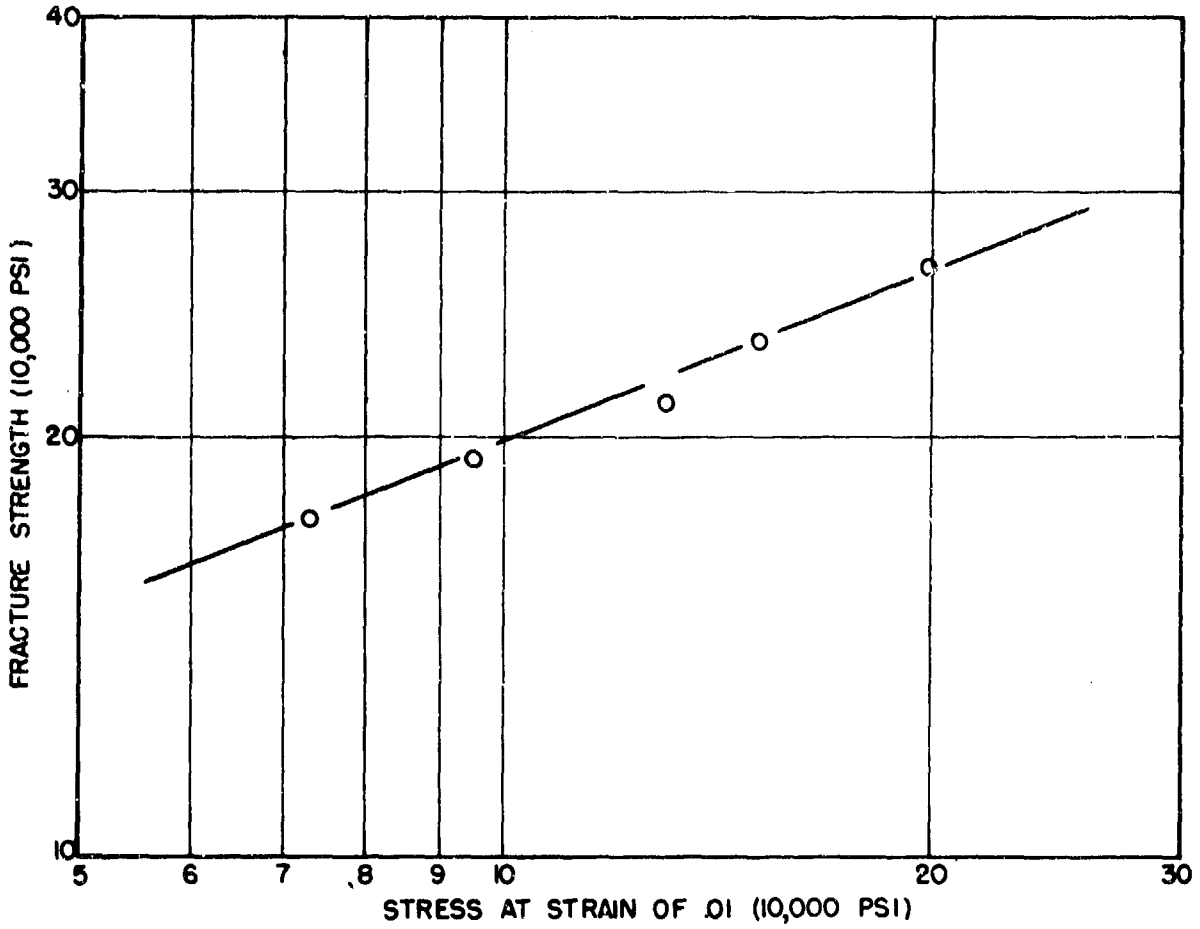


FIGURE 33

RELATIONS BETWEEN FRACTURE STRENGTH AND FLOW STRENGTH OF TEMPERED MARTENSITIC STEEL. (AFTER HOLLOMON[163])

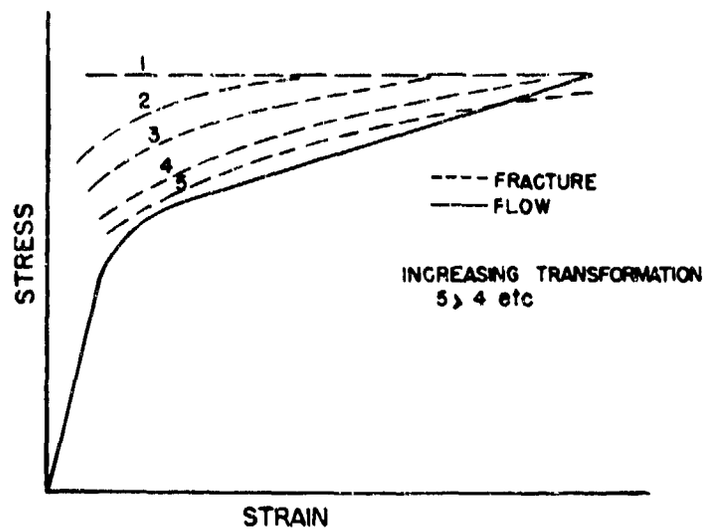


FIGURE 34

EFFECT OF TEMPER BRITTLENESS PRECIPITATE ON FRACTURE-STRESS CURVES (SCHEMATIC) (AFTER HOLLOMON[129])

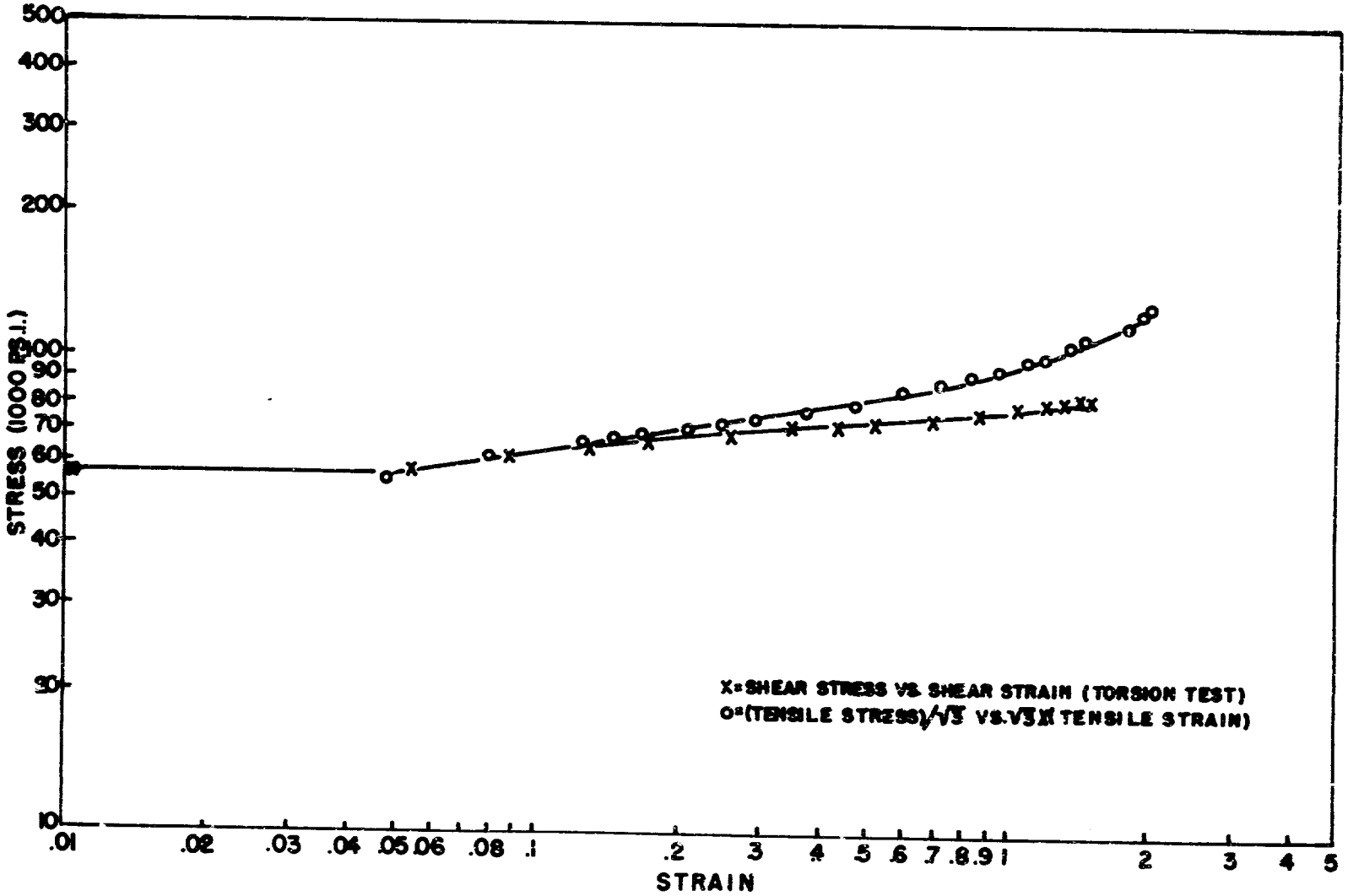


FIGURE 35

COMPARISON OF LOGARITHMIC FLOW-STRESS CURVES IN TENSION & IN SIMPLE SHEAR (TORSION) FOR A MARTENSITIC STEEL (TEMPERED)

(AFTER ZENER & HOLLOMON[177])

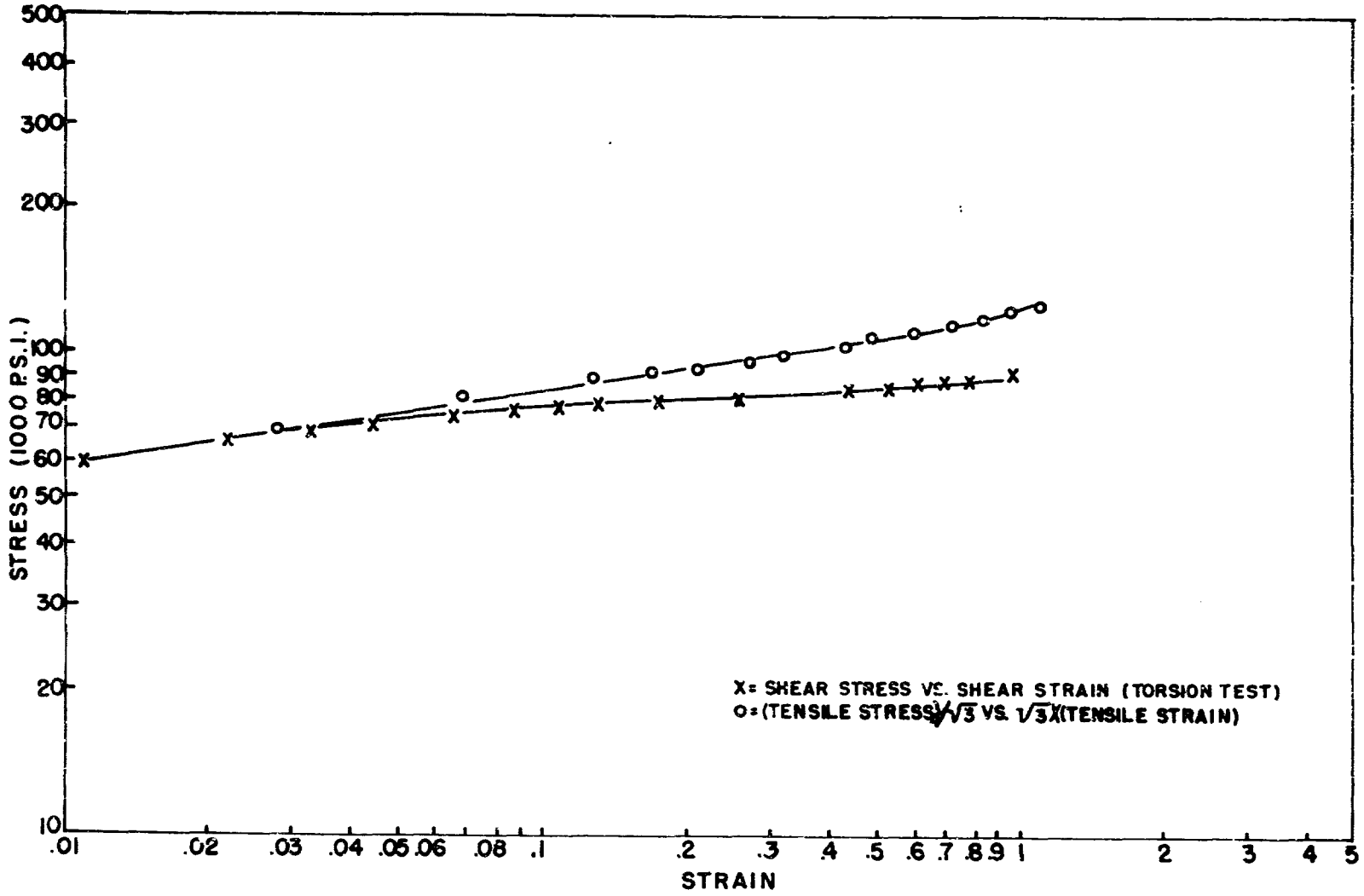


FIGURE 36

COMPARISON OF LOGARITHMIC FLOW-STRESS CURVES IN TENSION & IN SIMPLE SHEAR (TORSION) FOR A PEARLITIC STEEL.

(AFTER ZENER & HOLLOMON[177])

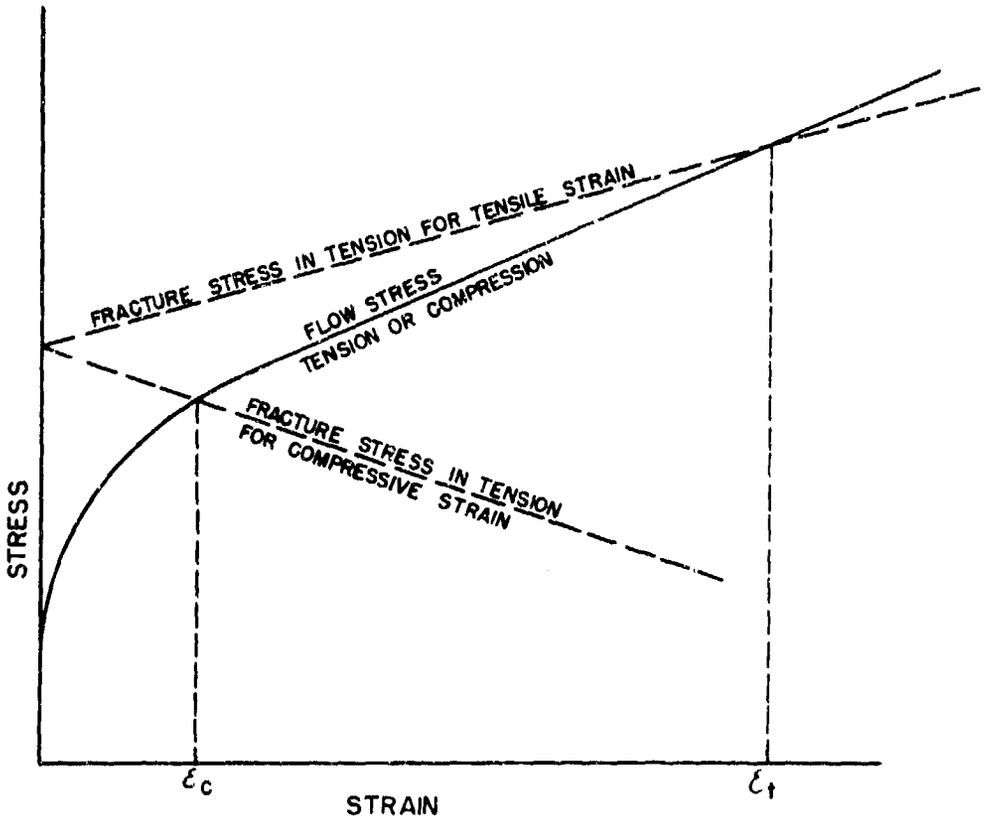


FIGURE 37

MECHANISM OF THE DECREASE OF DUCTILITY
IN TENSION WITH PRIOR COMPRESSIVE STRAIN

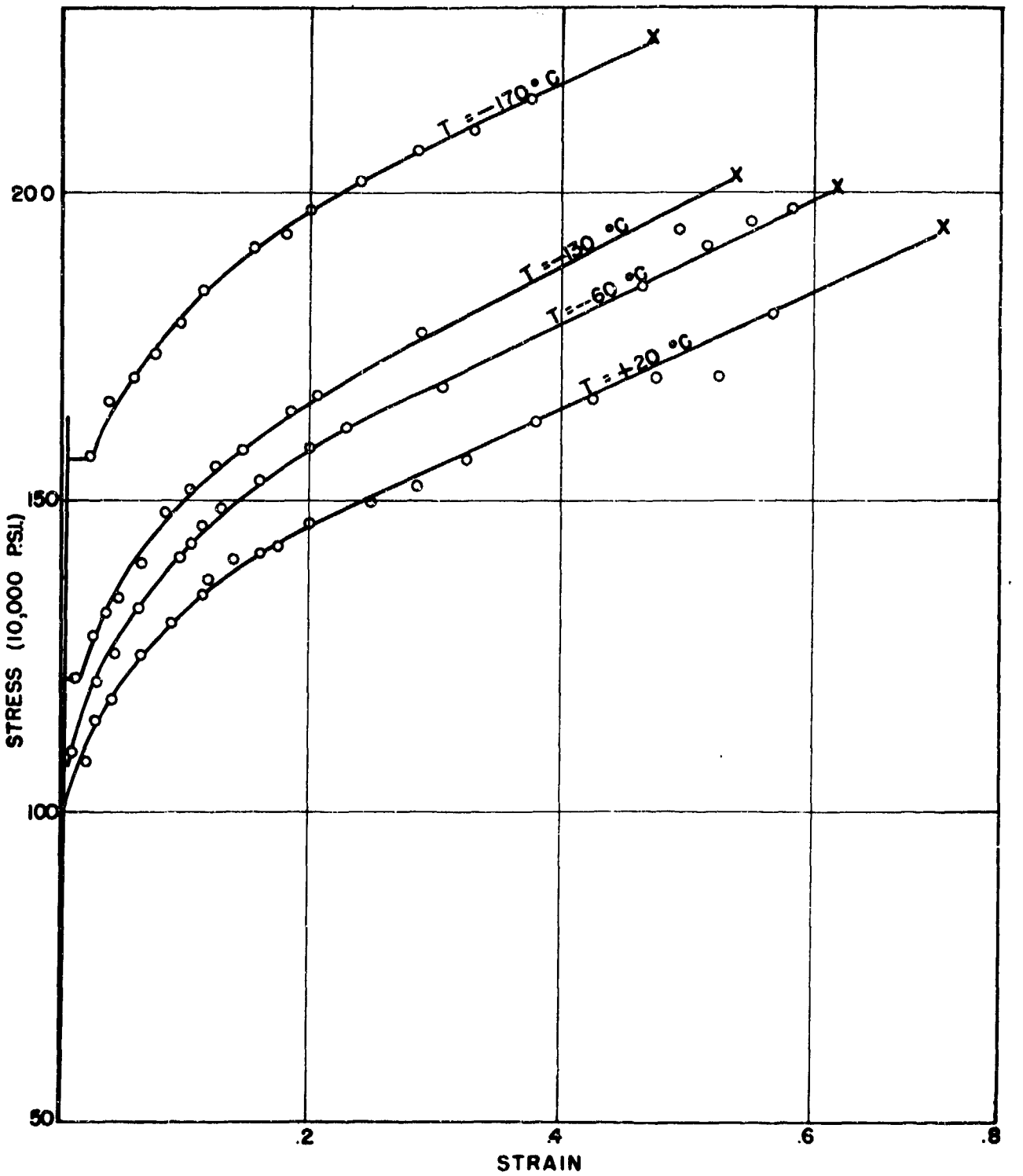


FIGURE 38

EFFECT OF DECREASING TEMPERATURE UPON FLOW STRESS CURVES OF A PEARLITIC STEEL. (AFTER ZENER & HOLLOMON[182])

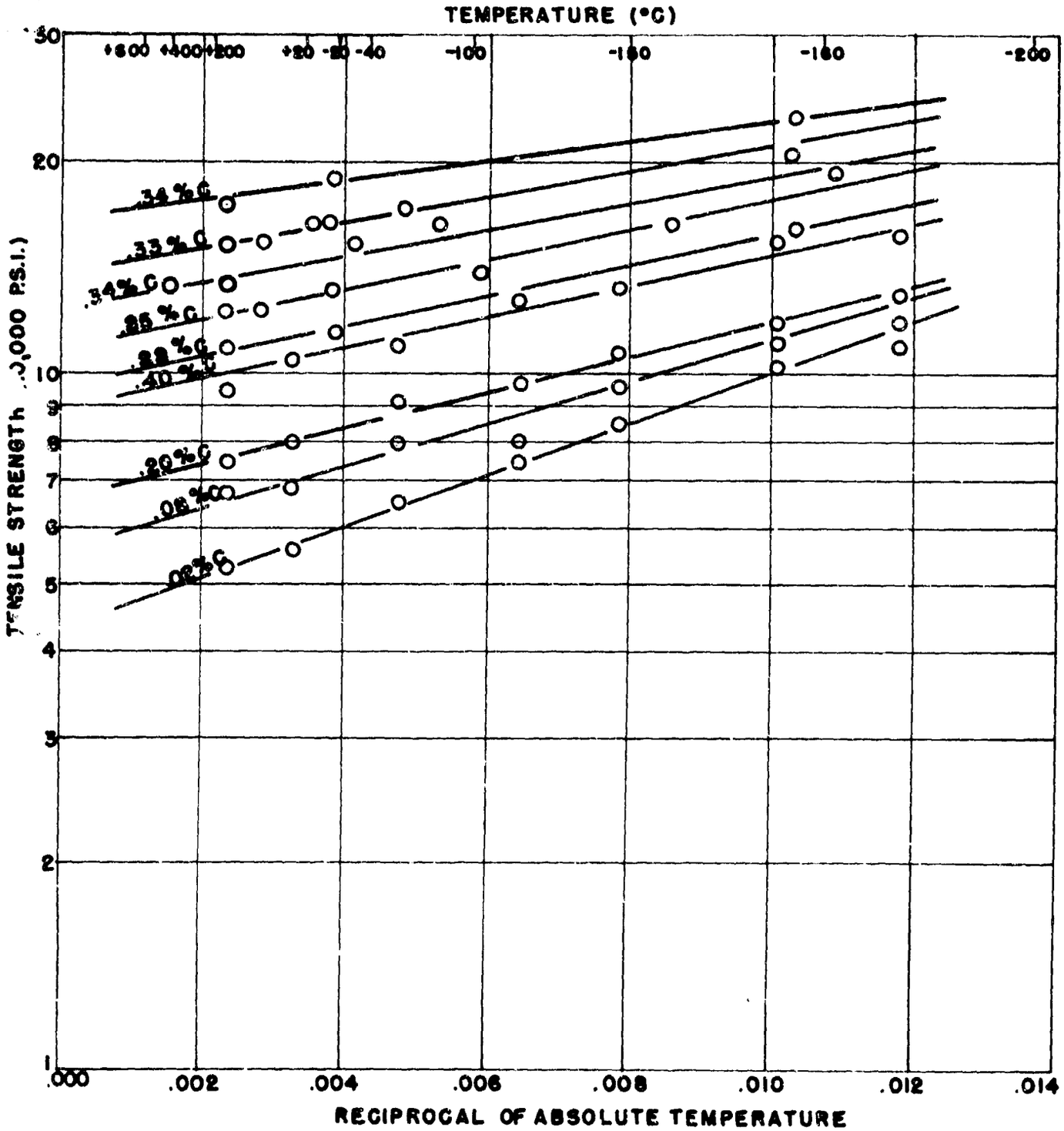


FIGURE 39

RELATION BETWEEN TENSILE STRENGTH & LOW TEMPERATURE FOR SEVERAL PEARLITIC STEELS. (AFTER HOLLOMON[189])

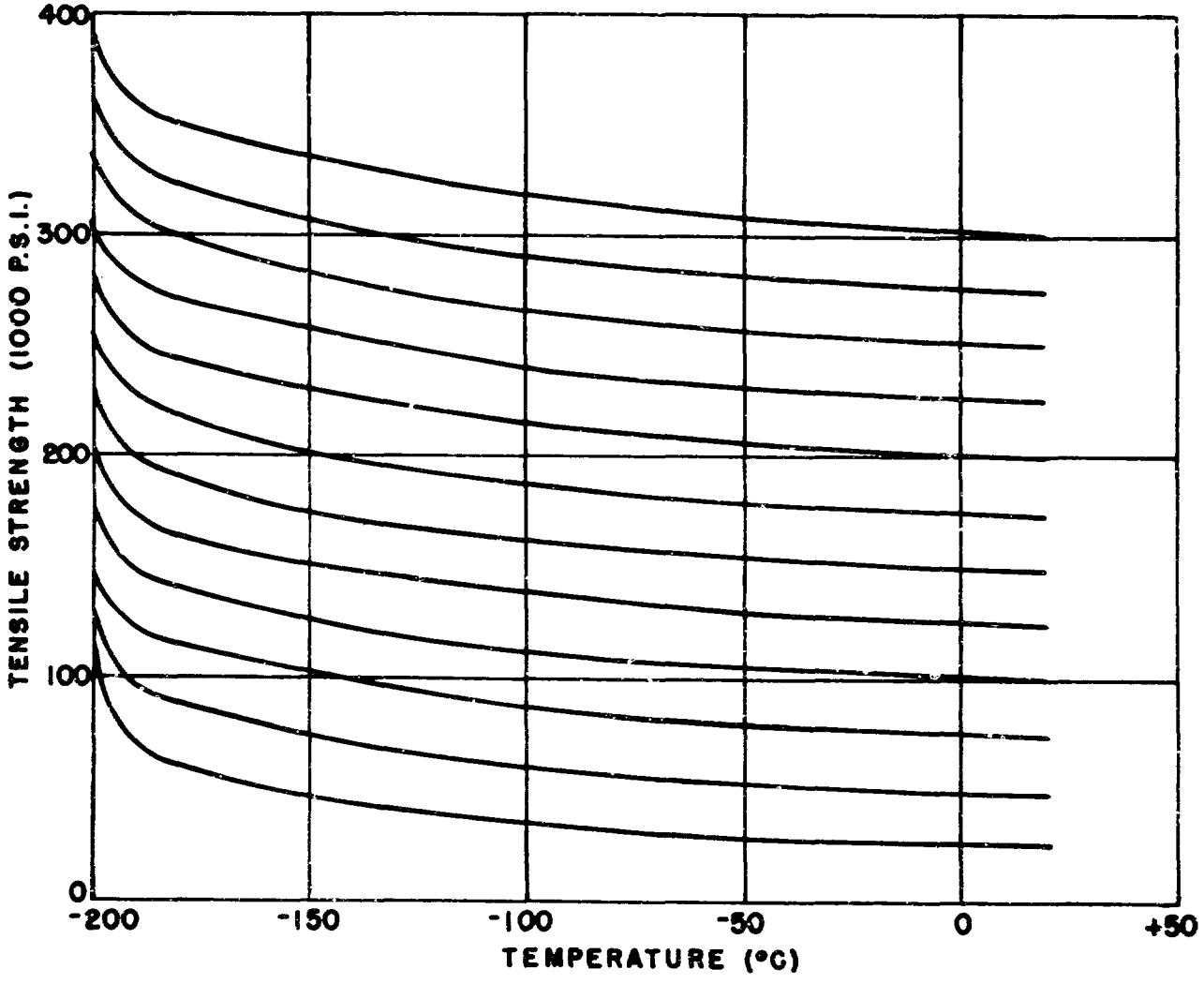


FIGURE 40
EFFECT OF LOW TEMPERATURE ON THE TENSILE STRENGTH OF PEARLITIC STEEL AT VARIOUS STRENGTH LEVELS.

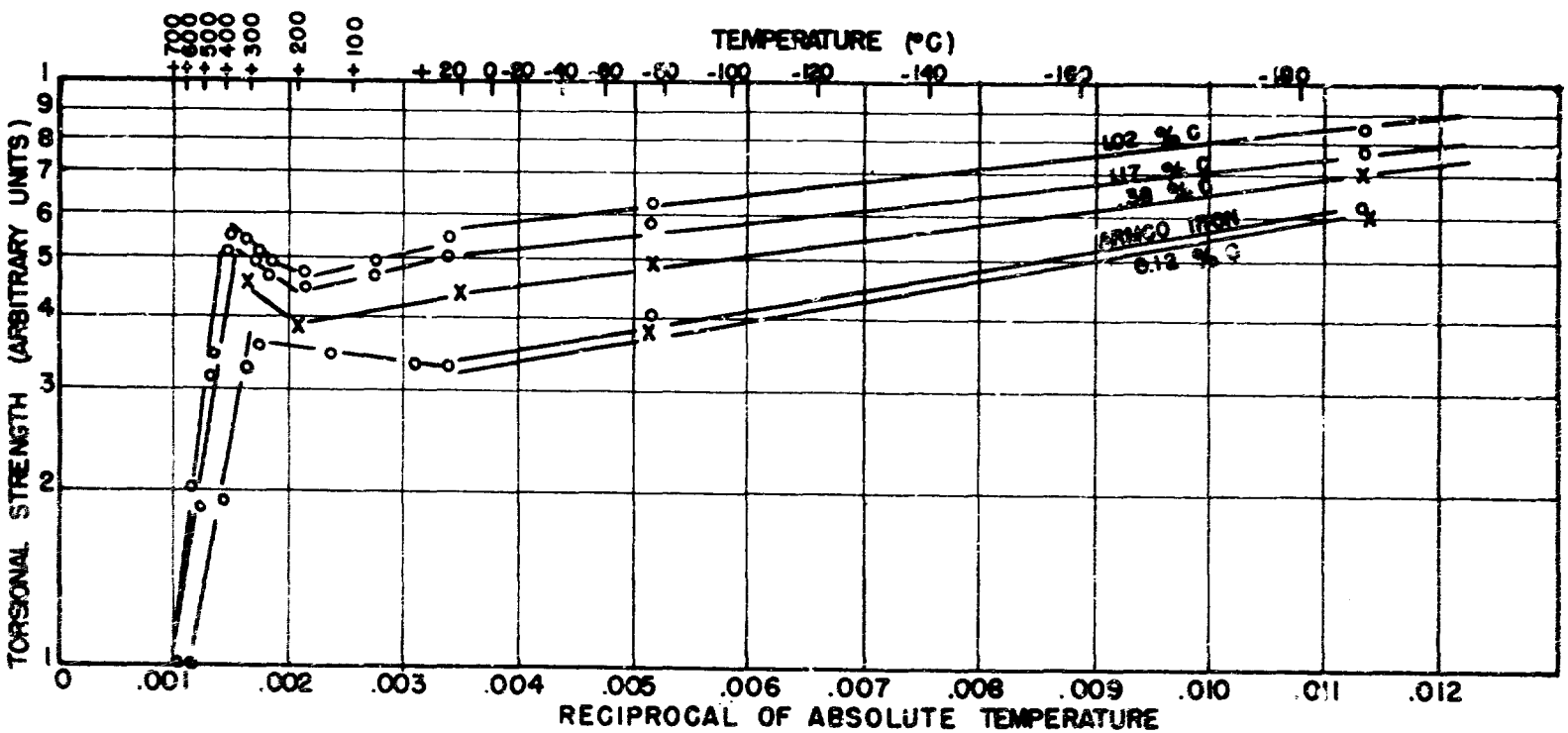


FIGURE 41

EFFECT OF TEMPERATURE UPON TORSIONAL STRENGTH OF SEVERAL PEARLITIC STEELS (FROM DATA OF LEITER[186])

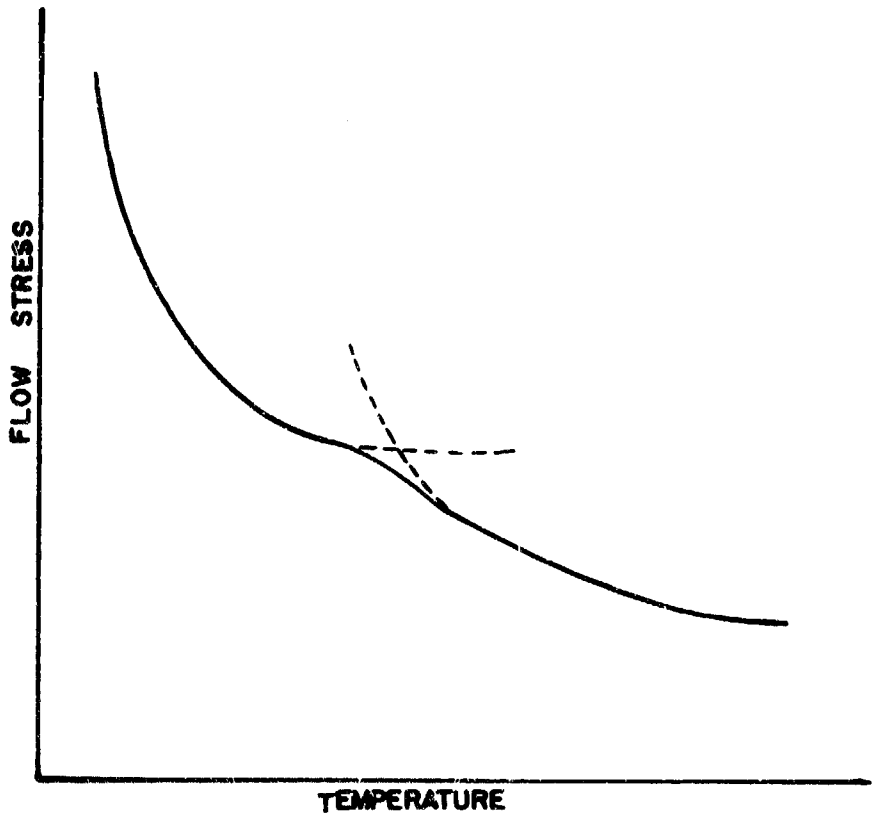


FIGURE 42

VARIATION OF FLOW STRESS AT A SMALL STRAIN WITH TEMPERATURE (SCHEMATIC).

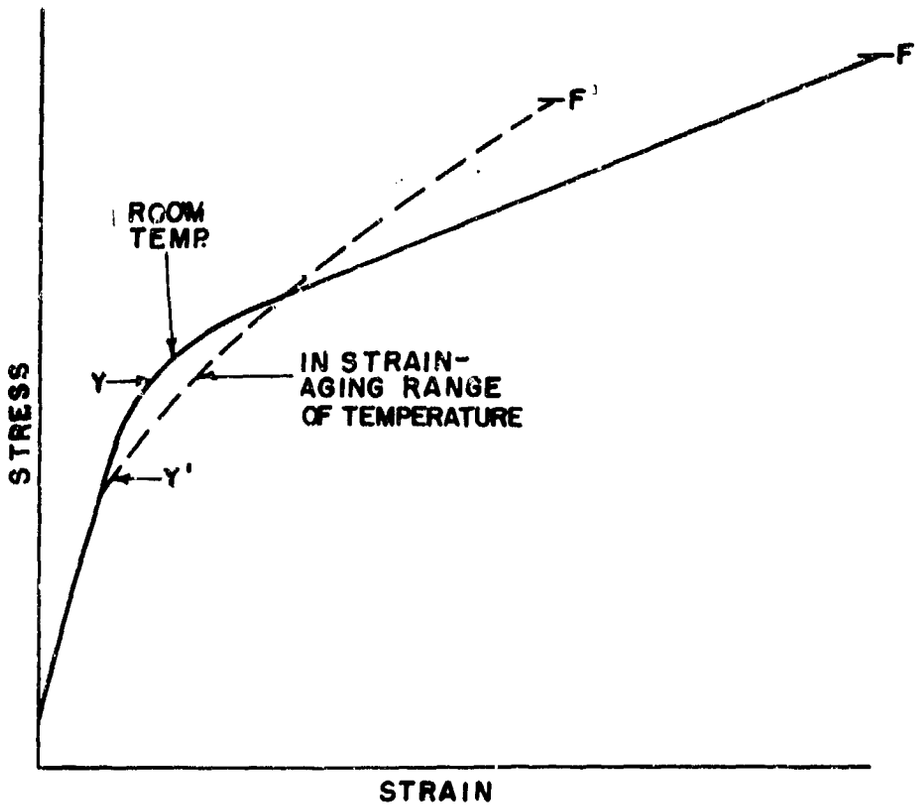


FIGURE 43

EFFECT OF STRAIN AGING ON THE FLOW STRESS CURVE (SCHEMATIC)

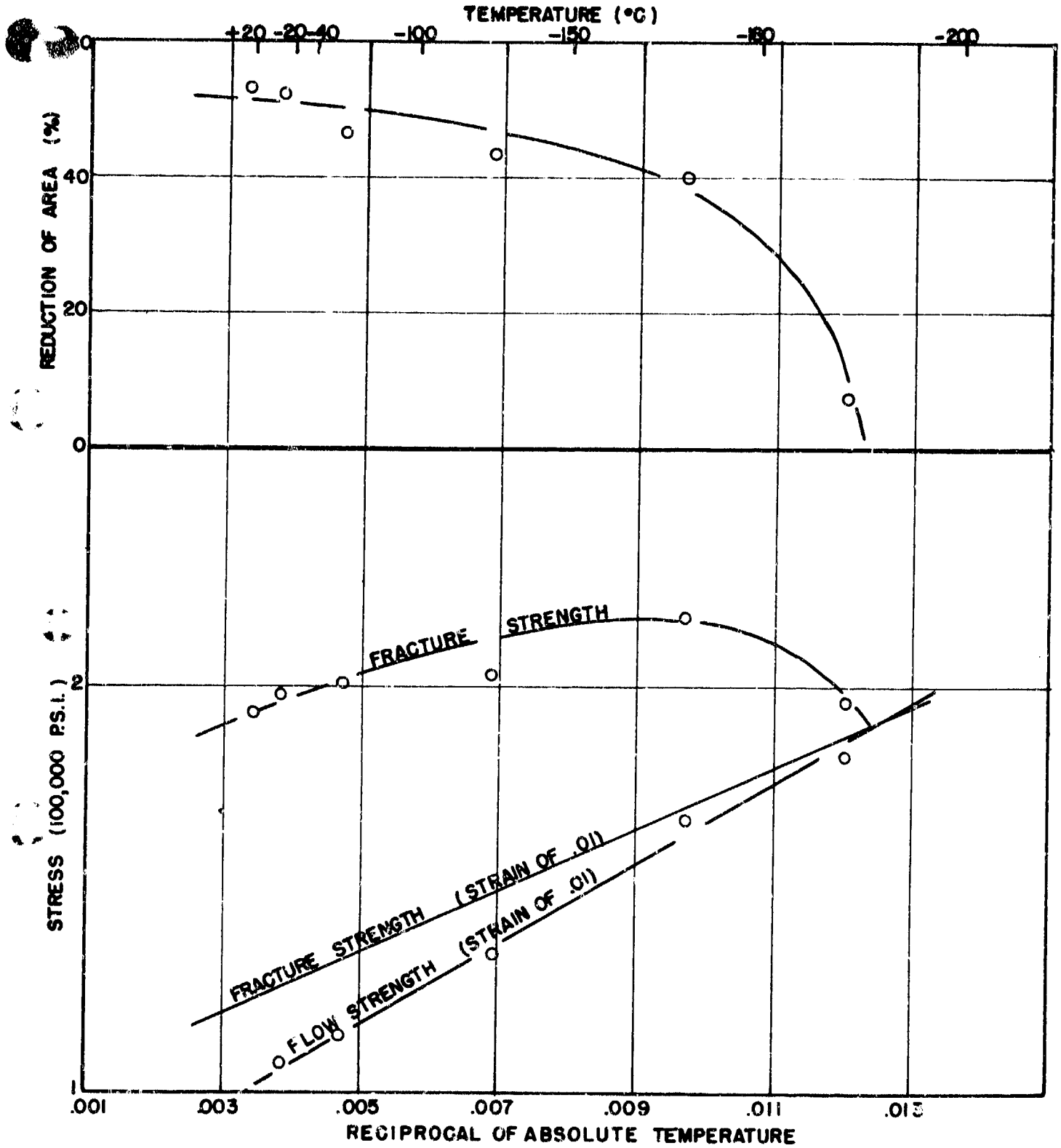


FIGURE 44

EFFECT OF LOW TEMPERATURE ON THE TENSILE PROPERTIES OF A PEARLITIC STEEL (AFTER HOLLOMON AND ZENER[166,168])

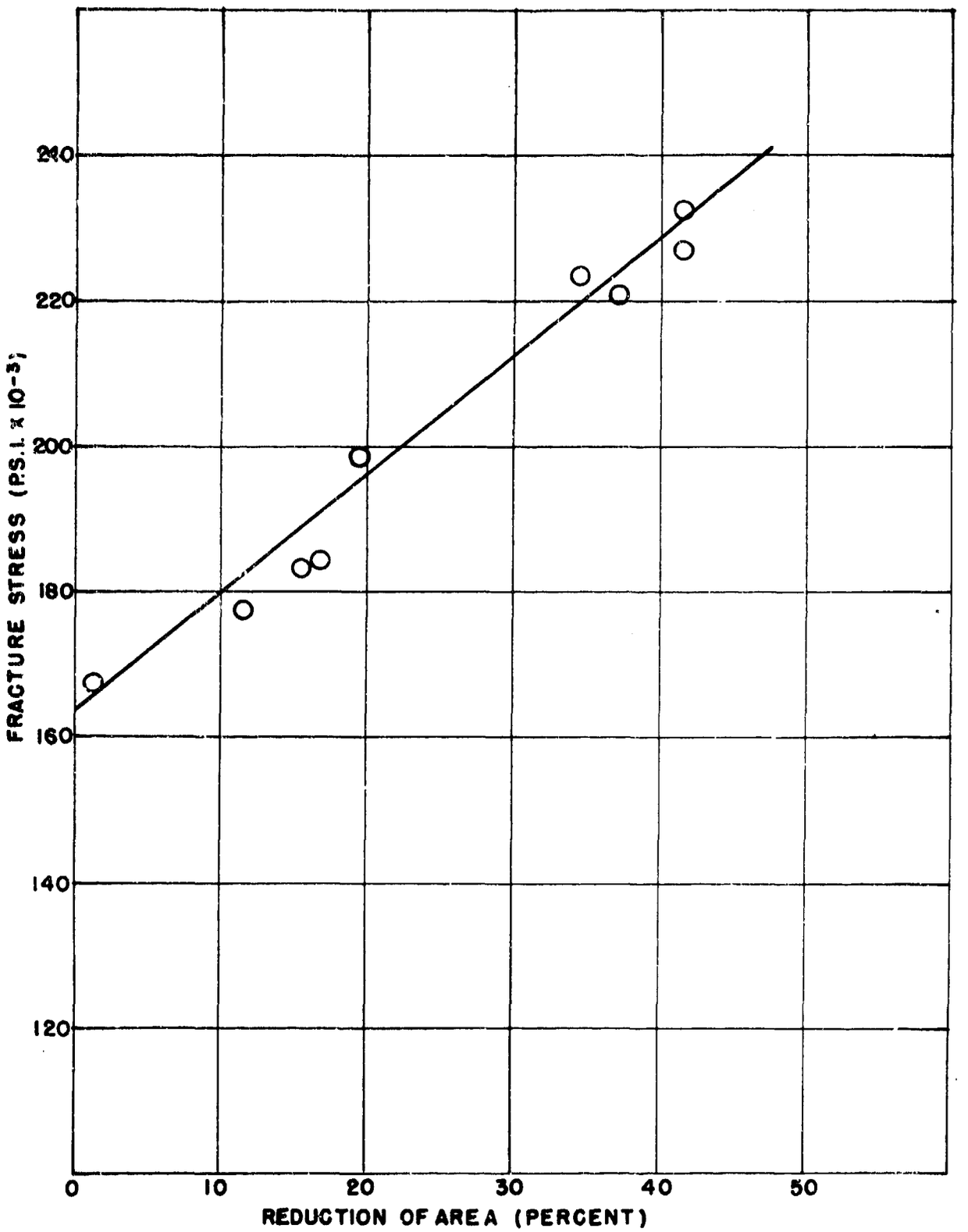


FIGURE 45
EFFECT OF PRIOR DEFORMATION ON
FRACTURE STRESS (AT -310 °F)
(PEARLITIC STEEL)
(AFTER HOLLOMON[9])

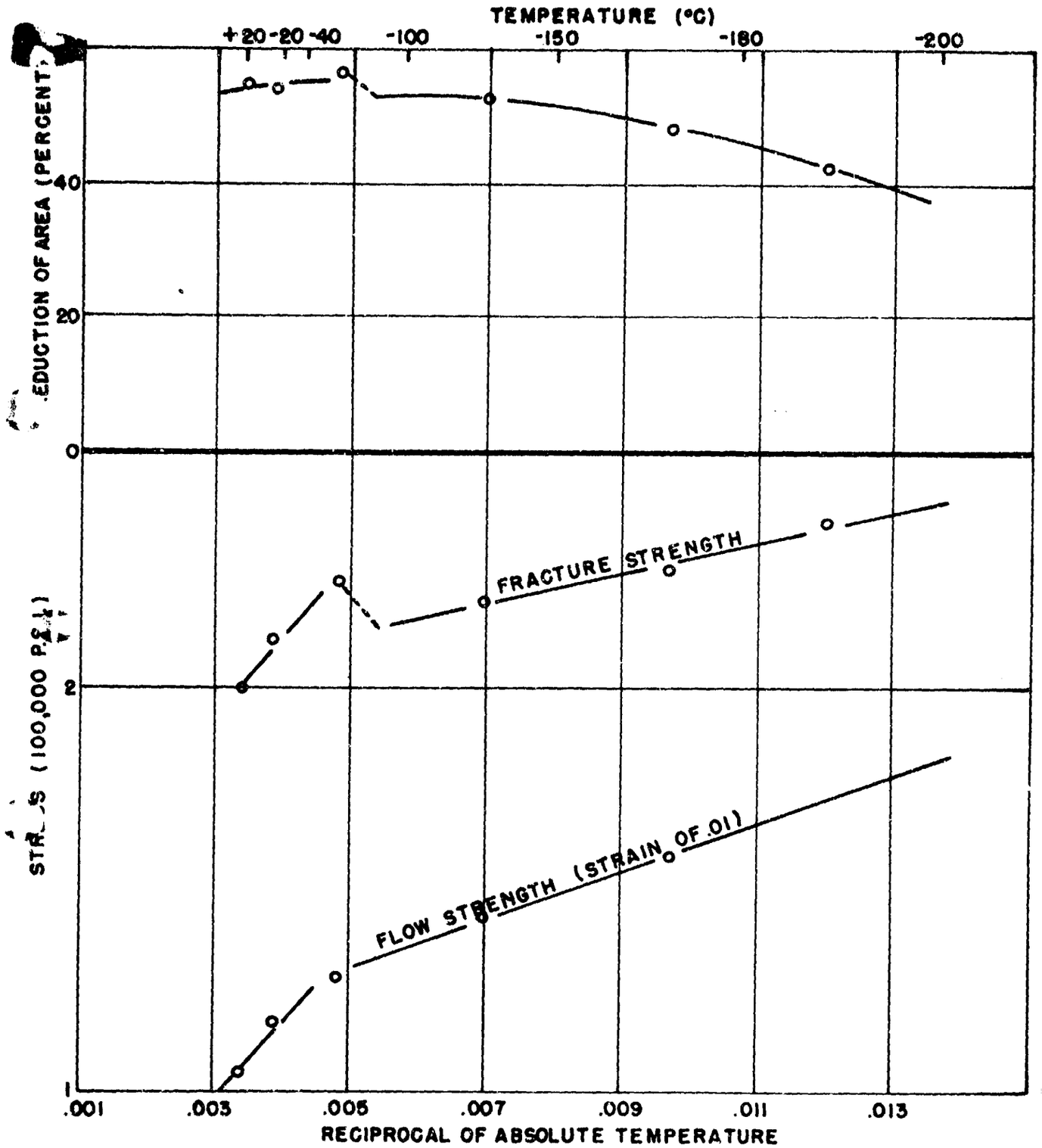


FIGURE 46

EFFECT OF LOW TEMPERATURE ON THE TENSILE PROPERTIES OF A TEMPERED MARTENSITIC STEEL (AFTER HOLLOMON & ZENER[180, 182])

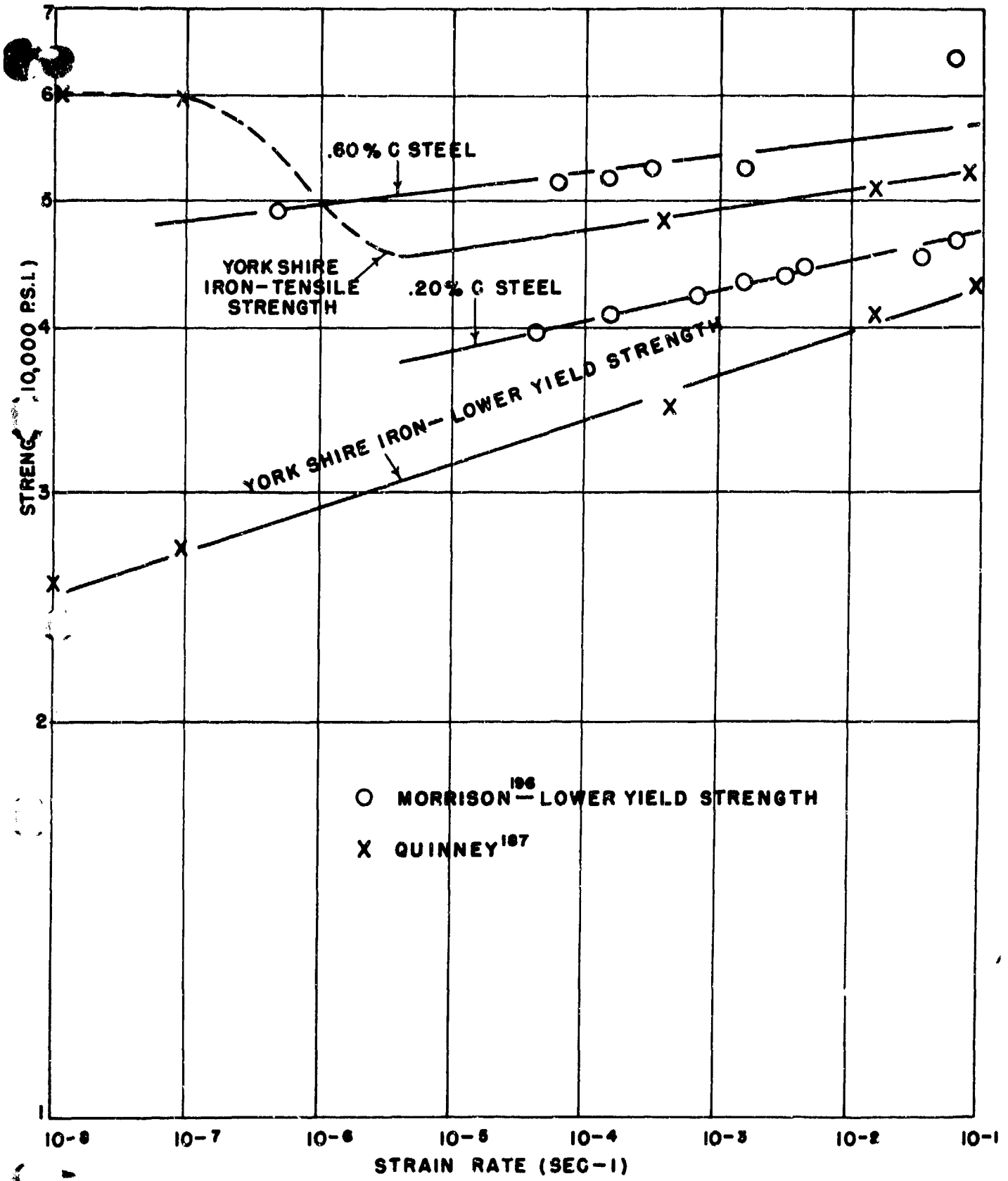


FIGURE 47

EFFECT OF STRAIN RATE ON THE YIELD & TENSILE STRENGTHS OF SEVERAL STEELS

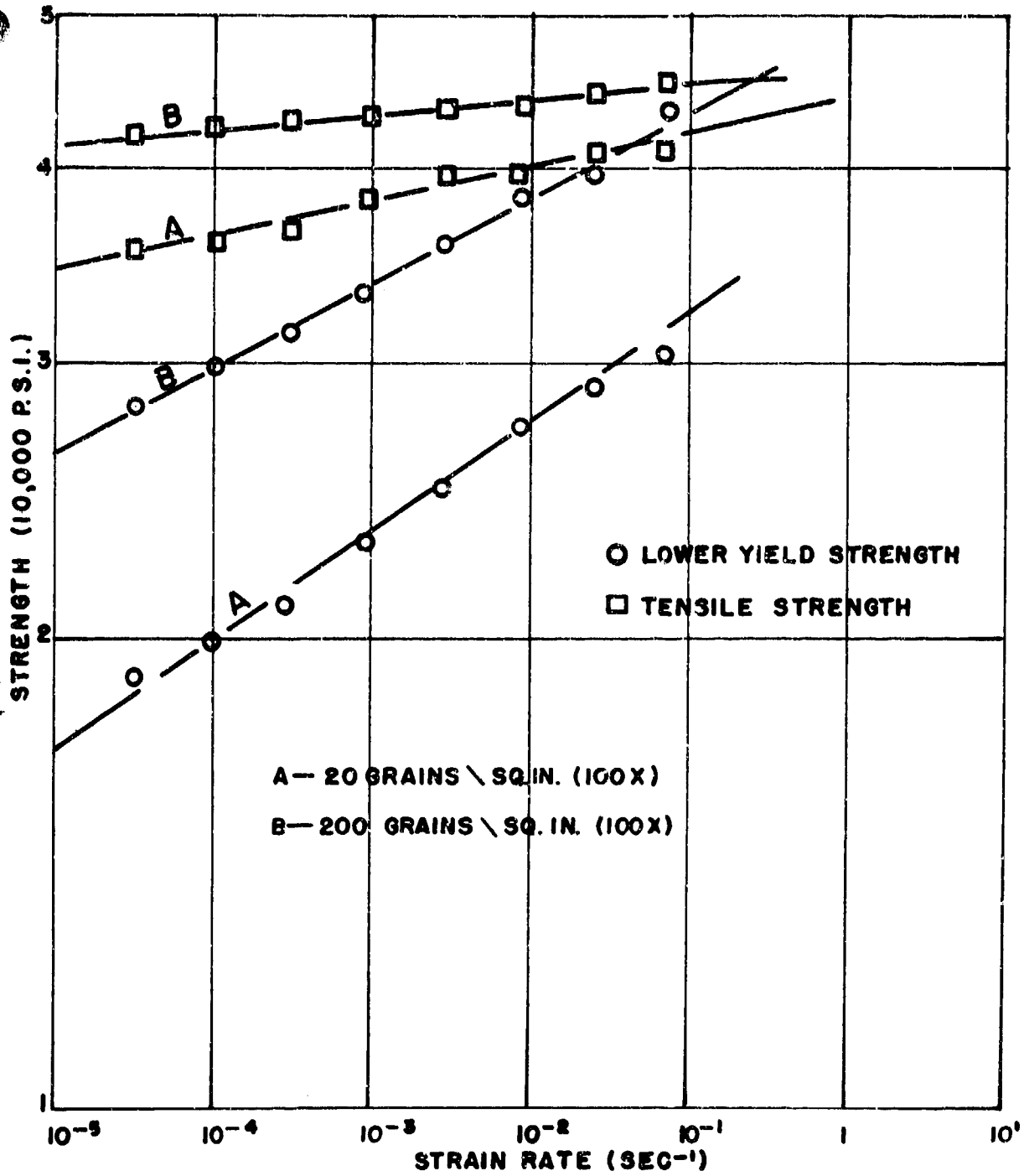


FIGURE 48

EFFECT OF STRAIN RATE ON THE YIELD & TENSILE STRENGTHS OF SEVERAL LOW CARBON STEELS (AFTER WINLOCK AND LEITER[197])

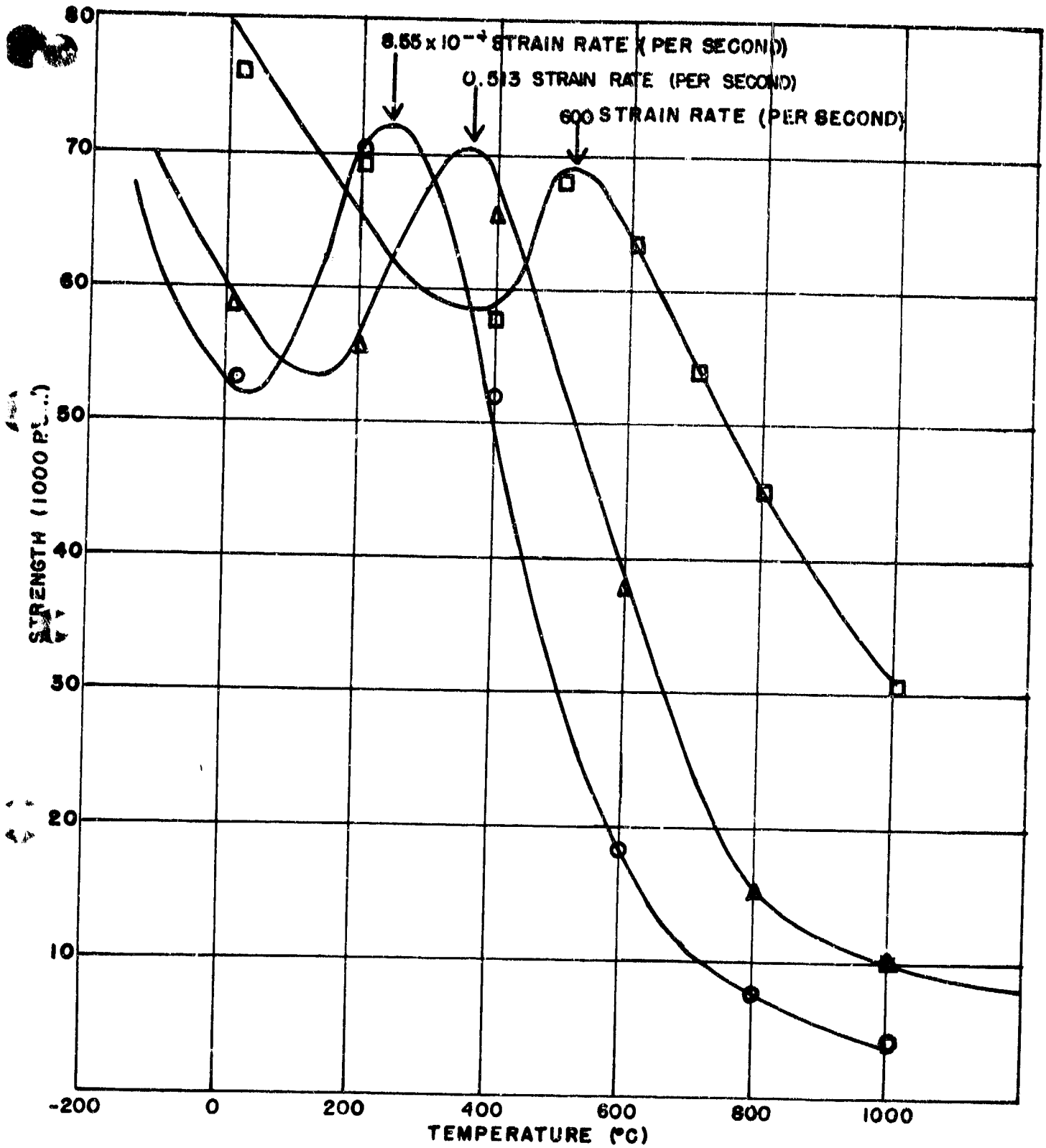


FIGURE 49

EFFECT OF TEMPERATURE & STRAIN RATE ON THE TENSILE STRENGTH OF STEEL.

(AFTER NADAI & MANJOINE[195])

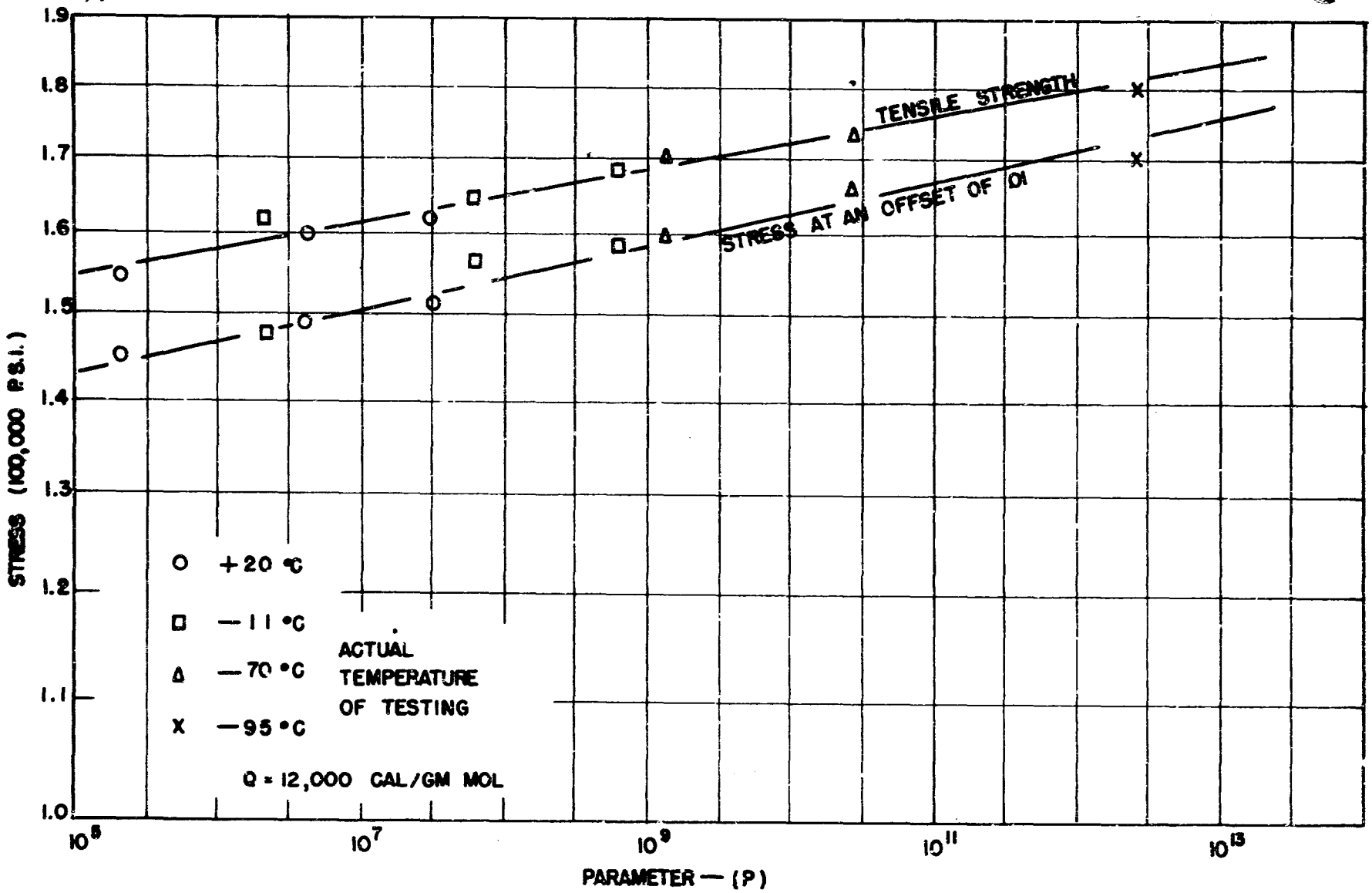


FIGURE 50

DEMONSTRATION OF EQUIVALENCE OF STRAIN RATE & OF TEMPERATURE FOR A TYPICAL STEEL (CASTING)*
(AFTER HOLLOMON & ZENER[16])

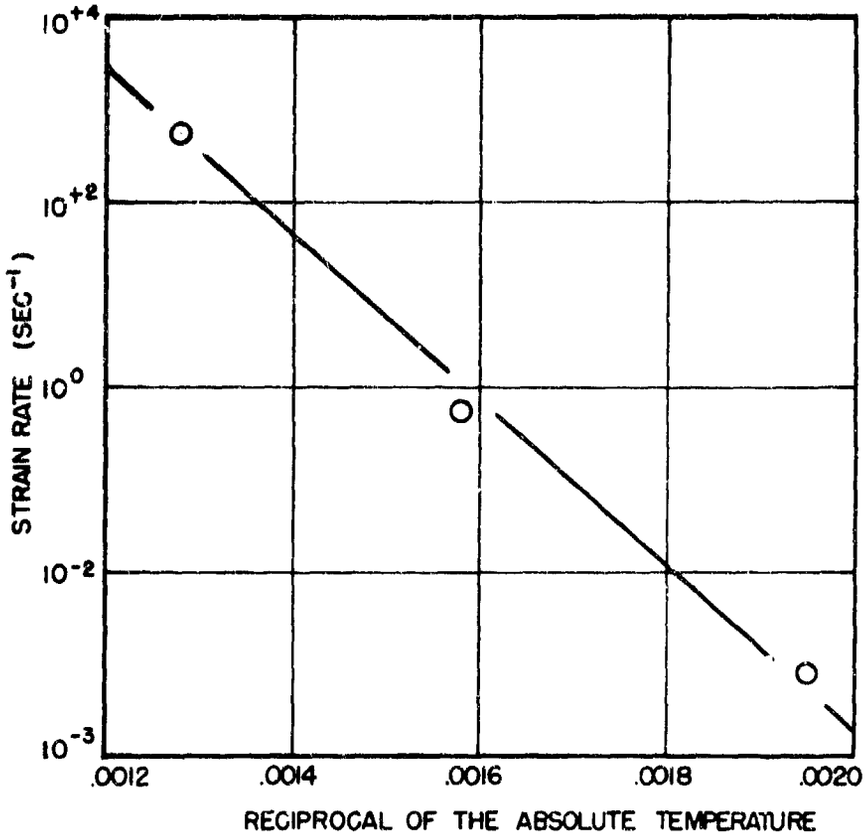


FIGURE 51

RELATION BETWEEN TEMPERATURE AND STRAIN RATE FOR MAXIMUM STRAIN AGING (FROM FIGURE[49])

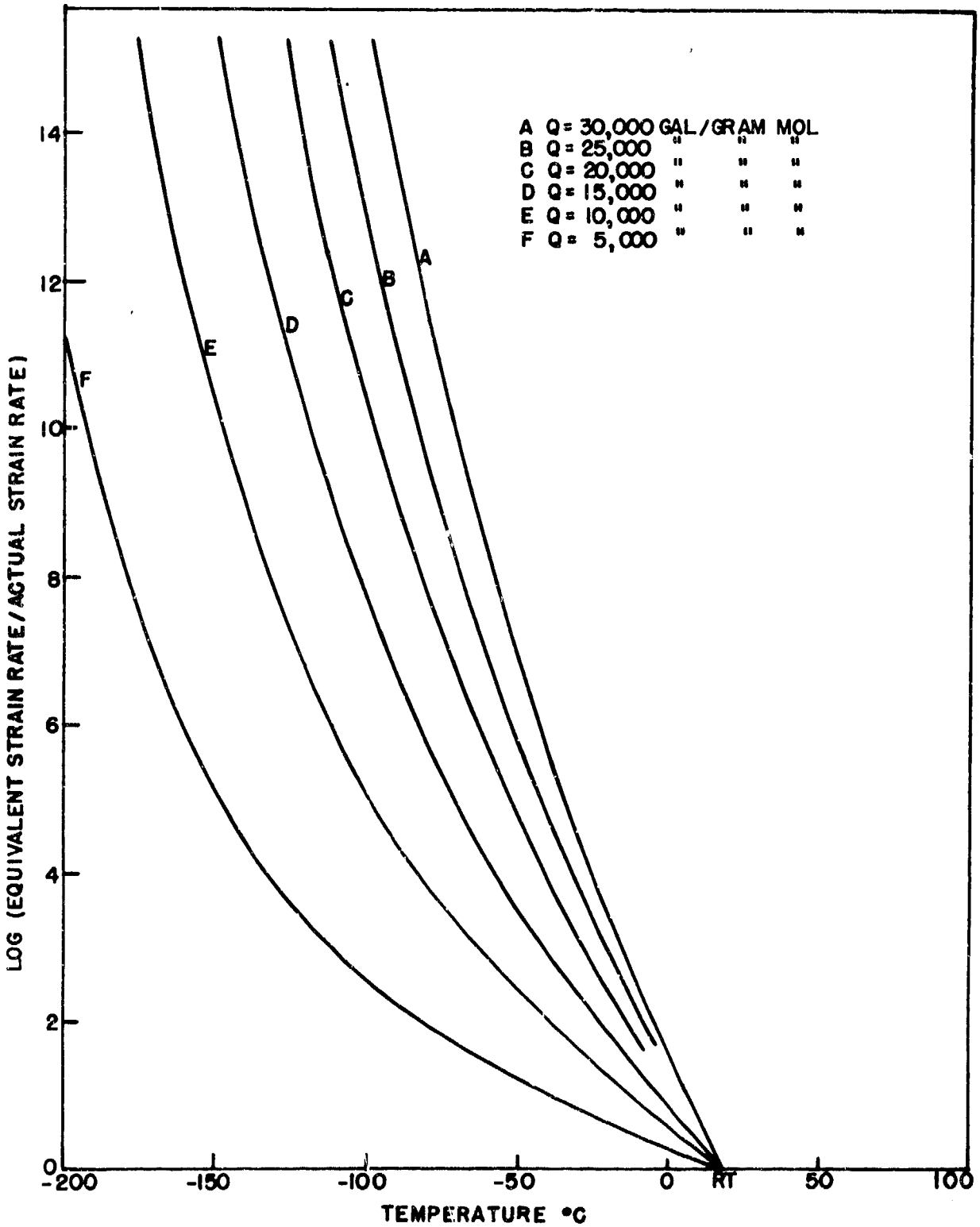


FIGURE 52

EQUIVALENCE OF A DECREASE OF TEMPERATURE TO A RISE IN RATE OF STRAIN. (AFTER ZENER & HOLLOMON[161])

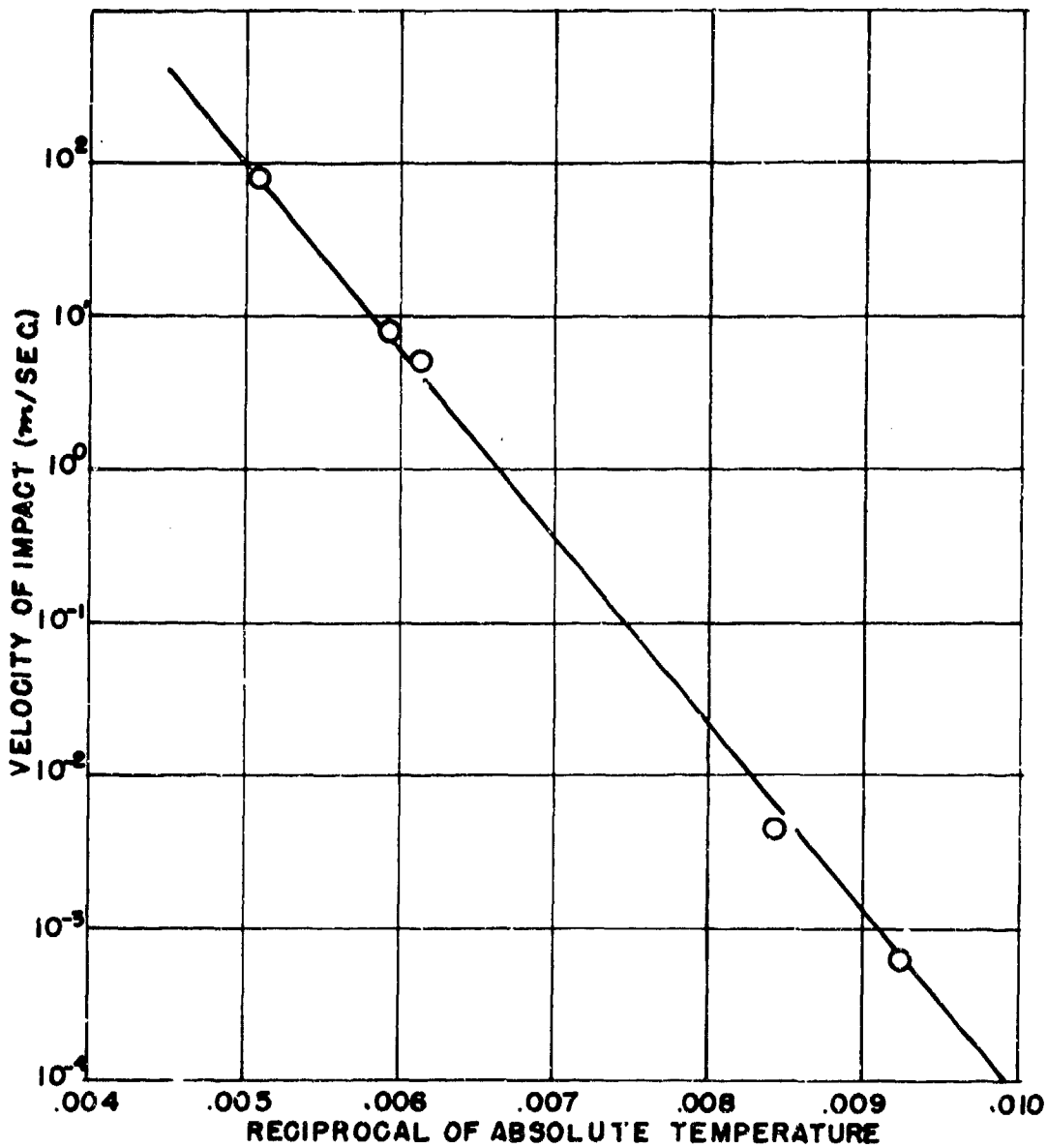


FIGURE 53

EFFECT OF IMPACT VELOCITY ON THE TEMPERATURE OF BRITTLE FAILURE (UNNOTCHED SPECIMENS) (AFTER WITMAN & STEPANOFF[200])

Suggested Reading

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"Elements of Engineering Mechanics" (In German) Julius Springer, Berlin (1909).

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"Plasticity", McGraw-Hill, New York, 1931.

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"Plastic Flow and Rupture of Metals", Trans. Am. Soc. Metals, 33 (1944) 163-235.

C. Zener and J. H. Hollomon:
"Problems in Plastic Deformation of Metals", Submitted to J. Applied Phys., 1945.

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IV. MECHANICAL PROPERTIES

The interpretation of most of the results of the standard mechanical tests (the mechanical properties) can be based upon an understanding of the effects of variables on the flow and fracture curves discussed in the preceding chapter. When the variables of the tests are known, their effects on the flow and fracture curves can be determined and the results of the tests predicted. Better information, particularly with respect to fracture stress, will permit more accurate prediction and correlation of the results of the mechanical tests.

It is almost always necessary to know the strength of steel to be used in heat-treated parts, in order that the stress it can withstand without excessive plastic flow can be estimated. Also, it is frequently desired to determine the amount of deformation the metal can withstand without local contraction ("necking"). The strain to fracture is often used as a relative measure of steel quality. These properties of strength and ductility are generally measured in the tensile test. This test, however, gives little indication of the shape of the fracture stress curve and hence, little information on the behavior of steels under combined stresses or upon repeated deformation. The energy required to break a notched bar (the "toughness") and the variation of this energy with testing temperature afford a relative measurement of the separation between the flow and fracture curves. Fatigue studies reflect the relative ability of metals to resist fracture under conditions of repetitive loading. Sometimes heat-treated steel parts must sustain loads over long periods*; the required property of creep resistance can be qualitatively related to the flow curves.

Tensile and Related Tests

The tensile test is most generally used to determine the strength of metals and to assess their suitability for specific engineering applications. It must be remembered, however, that the results of this test are only relative, since the test is performed under conditions generally widely different from those occurring in service. Further, if a steel to be used under complex loading conditions has adequate strength and adequate ductility in simple tension, it may nevertheless fail brittlely (fracture without deformation) in service. Also, the service of heat-treated parts may be such that transformations occur during their operation and change the properties of the steel. For example, steels to be used in boilers and refineries have been found to fail because of temper embrittlement occurring at operating temperatures(201-203). Results of tensile tests must be used with discretion, for these tests are generally performed at room temperature, slow rates of strain, and under the conditions of an essentially

*More generally, special alloys are used if resistance to creep is required at high temperatures.

uniform uniaxial stress. The effects of temperature or strain rate may often be taken into account in applying the results of tensile tests to mechanical design, but the effect of stress distribution on the plastic behavior and fracture of steels is less easy to estimate. Furthermore, the effects of all of these variables are interrelated.

Description of Test

The tensile test is usually performed on flat or round specimens of standard lengths and cross-sections. Flat specimens are usually used with strip or sheet material from which round specimens cannot be obtained. Standard tensile tests are generally carried out at about room temperature and at rates of strain of about 10^{-4} sec.$^{-1}$. During tensile deformation, the cross-sectional area of the specimen and the rate of strain-hardening decrease. After sufficient deformation the stress may increase sufficiently rapidly, due to the decrease in cross-section, to cause plastic flow with a decreasing load. The deformation then becomes localized, and the specimen "necks" as it continues to deform under the action of the decreasing load. Thus, in the tensile test, the load first increases and then decreases, as indicated by the schematic load-elongation curve of Figure 54.

The condition for the initiation of localized deformation (necking) in simple tension may thus be expressed in the following form:

$$dL = 0 \qquad (4\text{-}1)$$

where L is the load. Since the load is equal to the stress, σ, times the area, A, equation (4-1) becomes:

$$d(A\sigma) = Ad\sigma + \sigma dA = 0. \qquad (4\text{-}2)$$

However, during plastic deformation the volume remains essentially constant and:

$$d(Al) = Adl + ldA = 0 \qquad (4\text{-}3)$$

where l is the length of a small section of the specimen. Combining equations (4-2) and (4-3), the following results:

$$\frac{d\sigma}{d\varepsilon} = \sigma. \qquad (4\text{-}4)$$

"Necking" will occur in simple tension tests, therefore, at the strain at which the slope of the flow-stress curve becomes equal to the stress at that strain(204). If, however, equation (3-5) adequately describes the relation between stress and strain in the region of strain in which necking occurs, then:

$$\frac{\frac{d\sigma}{\sigma}}{\frac{d\varepsilon}{\varepsilon}} = m \qquad (4\text{-}5)$$

and upon substitution in equation (4-4):

$$\varepsilon_m = m. \qquad (4\text{-}6)$$

Necking, then, will commence in simple tension at a strain (ε_m) that is exactly equal to the slope of the logarithmic flow-stress curve. This localized deformation frequently limits the amount of deformation a metal can suffer. In many forming operations, the metal "fails" when necking begins and the extension becomes localized. After localization of the deformation the metal fractures. Even though in most forming operations the stresses are more complex than simply tensile, the amounts of deformation that metals can withstand without necking can be compared by comparing the strains to maximum load in simple tension(163).

After necking commences, the longitudinal tensile stress in the deforming region of a round tensile specimen is no longer uniform and there exists a circumferential tension and a radial tension*. (The radial tension must be zero at the surface of the specimen.) Therefore, after necking begins, the stress in a tensile specimen is no longer uniform or uniaxial and the stress obtained by dividing the load by the instantaneous area is only an average stress. An attempt(205) has been made to analyze theoretically the stress distribution in the neck of a tensile specimen. This analysis indicates that for ordinary tensile tests of metals the maximum longitudinal tensile stress is only slightly greater than the average stress. The maximum tensile stress exists at the axis of the specimen, and, in cases where necking has proceeded sufficiently, causes the tensile fracture to begin at the center of the specimen(206). The presence of the radial and circumferential stresses may in some steels induce longitudinal rather than transverse fractures (as will be discussed later).

The parameters generally measured in the tensile test are yield strength, tensile strength, reduction of area before fracture, and elongation of a specified gage length before fracture. In general terms, the yield strength is a measure of the height of the flow-stress curve, or of the stress required for initial plastic flow. The tensile strength is a measure of a combination of the initial height of the flow-stress curve and the degree of strain-hardening. The per cent elongation before fracture is primarily a measure of the amount of uniform deformation that occurs before necking, but is affected slightly by the amount of local deformation (necking) that takes place. The reduction of area is a measure of the strain at which the flow curve intersects the fracture curve under the particular conditions of the tensile test. For metals of the same metallurgical structure, the reduction of area is an indirect measure of the separation of the flow and fracture curves.

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*In the case of flat specimens the deformation is not uniform after necking commences and the cross-section does not remain rectangular. The strain then cannot be measured in a practical manner. Measurements of the strain to fracture are almost meaningless with flat specimens if "necking" occurs before breaking.

Flow Characteristics

Yield Strength

The yield strength is generally measured by determining the stress necessary to produce a specified total strain (including both elastic and plastic) or by the stress necessary to produce a specified plastic strain. For simplicity, the strain is usually specified in terms of per cent elongation*. Frequently, for steels exhibiting a drop of load at yielding, the upper or lower yield stress is taken as the yield strength, and is called the yield point.

The yield strength is commonly determined at strain offsets (plastic strains) of .01%, .1%, or .2%*. When steels that yield inhomogeneously are tested in hydraulic machines with considerable inertia, the drop in load at yielding does not occur discontinuously. Under such conditions, the .01% offset sometimes intersects the upper yield point and loses its significance, for the upper yield point is dependent upon the surface finish of the specimen, the axiality of loading, etc. In flow stress curves obtained by tests in other machines, the .01 and the .1% offset often are less than the lower yield strain and measure the lower yield stress. This lower yield point depends not only upon the gross microstructure of the steel, but also upon the amount of nitride or carbide that is present (perhaps precipitated at the grain boundaries). Increased nitrogen contents may, for example, raise the lower yield strength and increase the lower yield strain. For steels of high yield strength, the .1 and .2% offsets generally intersect the rising portion of the flow stress curve. The yield strength as determined at these offsets, therefore, depends primarily upon the number, size, distribution of sizes, and perhaps type of carbide.

There is little fundamental information relating the yield strength and the structure, and direct correlations are probably not feasible, since, as discussed in the preceding chapters, direct quantitative measurements of structures are virtually impossible. Relations are generally established between the strength and the treatment necessary to produce the structure, rather than between the strength and the structure directly. Since in the case of tempered martensitic steels all other properties are similar when the strengths (and carbon contents) are the same, such correlations are of extreme usefulness. The relations between strength and treatment (tempering) may, of course, depend upon composition. The variation of strength (as measured by hardness) with tempering treatment and composition is discussed in the chapter on temperability.

Tensile Strength

The tensile strength is determined by dividing the maximum load reached in the tensile test by the original area. The tensile strength is

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*For small strains the per cent elongation is equal to the strain multiplied by 100.

frequently a more satisfactory measure of the height of the flow-stress curve than is the yield strength, because of the variations in strength that arise from inhomogeneous deformation and from residual stresses introduced by quenching or by other forms of prior plastic deformation. It must be remembered however, that unlike the yield strength measured at small strains, the tensile strength is affected by changes in strain-hardening. By definition the tensile strength (S_o) is given by:

$$S_o = \frac{\sigma_m}{A_o} Am \qquad (4\text{-}7)$$

where σ_m is the stress at the maximum load, Am, the area at the maximum load and A_o the original area. Further, from equation (3-5):

$$\frac{\sigma_m}{\sigma_y} = \left(\frac{\varepsilon_m}{\varepsilon_y}\right)^{\varepsilon_m} \qquad (4\text{-}8)$$

where σ_m and σ_y refer respectively to the stress at maximum load and to the flow stress at a small strain ε_y. On substitution:

$$S_o = \sigma_y \left(\frac{\varepsilon_m}{\varepsilon_y}\right)^{\varepsilon_m} \frac{A_m}{A_o}$$

and

$$\ln \frac{\sigma_y}{S_o} = \varepsilon_m\left(1 - \ln \frac{\varepsilon_m}{\varepsilon_y}\right). \qquad (4\text{-}9)$$

Thus, the ratio of the stress (on the flow-stress curve) at a given strain to the tensile strength depends only upon the strain-hardening exponent (ε_m). If the steel exhibits a lower yield stress and this is taken as the yield strength, the ratio of yield strength to tensile strength will be higher than for steels that yield homogeneously. Thus, tempered martensitic steels of moderate tensile strengths have higher yield-tensile ratios than do pearlitic or bainitic steels of the same tensile strength. This difference in the yield-tensile ratio is often used as a practical indication of the type of quenched structure of tempered steels. For tempered martensitic steels the yield-tensile ratio is surprisingly reproducible, not varying to any great extent from heat to heat or from steel to steel. Curves relating the yield-tensile ratio with tensile strength for steels of different structures are presented in Figure 55*. In Figure 56 the yield strength (either yield point or .1% offset) is plotted as a function of the tensile strength for quenched and tempered steels**.

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*The slight drop of the ratio at high strengths is due to quenching stresses that are not relieved at the low tempering temperatures used in obtaining these strengths. These residual stresses decrease the additional stress necessary to initiate flow.

**The lower alloy steels (S.A.E. 10 series) were certainly not quenched completely to martensite and should not have been included in the graphs.

These data are for steels ranging from about .25% carbon to .45%. Increasing the carbon content tends to raise the tensile strength for a given yield strength (Equation (4-9)) because the strain-hardening exponent increases with increasing carbon content.

Hardness*

The resistance to initial plastic flow is more frequently measured by hardness than by tensile tests. Hardness tests are generally performed by measuring the depth or diameter of the impression made by a specified indentor under a specified load for a specified time. In all cases plastic deformation occurs and the extent of the indentation is a measure of the amount of plastic deformation under the complex stress conditions of the hardness test. The extent of this deformation depends upon both the height of the flow-stress curve and the initial strain-hardening. The stress distribution in the hardness test is exceedingly complex but is constant for a specific test and hardness number. The hardness should be directly related to the tensile strength since the latter also depends upon both the yield strength and the strain-hardening. In general, the correlation of hardness and yield strength is poorer, for the yield strength may vary considerably due to changes in the initial yielding even when the stress at large strains is constant. An empirical relation between the tensile strength and the Brinell hardness of steel is presented as Figure 56. The relation between Brinell and Rockwell C hardnesses of steels is presented in Figure 57, while Figure 58 gives the approximate relation between yield strength and Rockwell C hardness for tempered martensitic steels.

Since the hardness may be rapidly determined and is very reproducible, it is the most widely used measure of "structure".

Per Cent Elongation

In the tensile test, per cent elongation of some specified gage length is affected more by the deformation that occurs before necking than by the localized deformation, except in cases where the uniform deformation is very small or where the gage length is comparable to the length over which necking occurs. The per cent elongation is, therefore, primarily determined by strain-hardening rather than by the relation between the yield and fracture strengths. As the strength increases, the per cent elongation decreases because the strain-hardening exponent and thus the amount of uniform deformation decrease. Because of the intimate relation between the per cent elongation and the amount of uniform elongation, the variation of the strain to tensile strength with carbon content and with strength level should be indicative of the variation of the per cent elongation. The variation of the uniform strain (strain-hardening exponent) with strength (stress at a strain of .01) is presented in Figure 28. The variation of the per cent elongation with tensile strength for steels of from .25 to .45% carbon is presented as Figure 59. These relations should be very

*Hardness tests are described and discussed in detail in several books (209-211).

nearly independent of structure for, as has already been pointed out, strain-hardening is virtually independent of structure. If fracture occurs before necking, the per cent elongation is primarily fixed by the separation between the flow and fracture curves, rather than by the strain-hardening exponent. Fracture before necking is indicative of very small separations, even at low strains, of the flow and fracture curves. In such cases, the per cent elongation and the reduction of area have the same significance.

Fracture Characteristics

Reduction of Area

The tensile test is terminated by fracture. The height of the fracture curve probably depends upon the size and distribution of inhomogeneities, such as carbides and inclusions, and its slope upon their shape. In the case of cast steels, shrinkage defects and porosity also limit the fracture stress. For steels of the tempered martensitic structure, the fracture stress rises less rapidly than the flow stress as the structure is made finer, and the reduction of area before fracture, therefore, decreases as the strength increases. If specimens from forged or rolled steels are taken parallel to the direction of forging (or rolling), the reduction of area will be higher than for transverse specimens. The "dirtier" the steel, the greater will the effect of the forging or rolling operation in increasing the reduction of area longitudinally and decreasing it transversely. Because of the sensitivity of the fracture stress to the presence of inclusions and other defects, the reduction of area varies considerably from heat to heat and even varies considerably for specimens taken from different portions of the same heat. The variation of the reduction of area within a heat of steel having a uniform microstructure has essentially the form of the Gaussian error curve. If used for process control, reduction of area measurements must, like most other data, be treated statistically.

The change of reduction of area with tensile strength is shown in Figure 60 for specimens taken from quenched and tempered bars of a number of (.25 - .45% carbon) steels. Clearly the reduction of area does not depend to an appreciable extent upon alloy content but varies significantly only with strength and steel quality. Variations from heat to heat of the same steel, or even within one heat of the same steel, are as great as the variations between steels of different compositions(208, 212). Increasing the carbon content of tempered martensitic steels decreases the fracture stress for a given strength level, and increases the slope of the flow-stress curves. For these reasons, the reduction of area decreases with increasing carbon content. The effect is small and is frequently impossible to distinguish because of heat to heat quality variations. Figure 61, however, shows data that indicate the trend.

Steels having structures other than tempered martensite generally have slightly lower reductions of area than tempered martensitic steels(161,

214)*. Non-martensitic products ordinarily lower the reduction of area only slightly. Similarly, the precipitate that is responsible for temper brittleness does not affect the flow curves or the reduction of area to any great extent. In extreme cases, the precipitate decreases the slope of the fracture stress curve sufficiently to cause a slight decrease in reduction of area, and a startling change in type of tensile fracture; the fracture changes from transverse to longitudinal. This change in type of fracture is more likely to occur if the specimen is taken parallel to the forging direction and if the structure of the steel is pearlitic. Longitudinal fracture occurs after a tensile specimen of a steel, having a fracture stress strongly dependent upon strain, necks sufficiently for the induced circumferential stress to cause longitudinal fracture. The longitudinal stress required for fracture perpendicular to the axis of the specimen increases with tensile deformation while the transverse tensile stress required for fracture decreases. The greater the precipitation, the smaller will be the strain before the specimen fractures longitudinally, and the lower will be the measured fracture stress. Pearlitic steels that have very low fracture stresses at small strains may fracture longitudinally even if not temper brittle.

Hydrogen may precipitate during heat treatment to cause defects such as "flakes"**. These "flakes" can initiate fracture, as do other defects, by concentrating the stress and thus lowering the reduction of area (171). Furthermore, in steels supersaturated with hydrogen, the gas may precipitate after a small amount of plastic deformation and cause relatively brittle fracture (hydrogen embrittlement). The exact mechanism of the effect is not known and there is considerable controversy concerning it. If the saturated specimens are deformed rapidly, little or no embrittlement is observed, presumably because there is insufficient time for hydrogen precipitation***. Thus, plastic deformation is necessary for the embrittlement to occur(216).

Bend Strength****

At high hardnesses stress concentrations arising from fillets, nicks, scratches, and non-axial loading may be sufficiently severe to cause brittle failure (fracture without deformation) in simple tensile tests. If, however, a small amount of plastic flow takes place the stress concentrations may be reduced and considerable deformation may occur before fracture. Minor variations in the fracture stress or in surface preparation, therefore, may be reflected in large changes in reduction of area at high strength

*An exception has been reported in the case of plain carbon steels, in which the bainitic structures have been stated to be superior(215). The difficulty of obtaining a completely martensitic structure on quenching of plain carbon steels in sections large enough even for test specimens should be borne in mind, however.

**These defects are also called "fish-eyes", "snow flakes", etc.

***These results were obtained in a research by S. A. Herres not yet published.

****Bend tests have been described in several publications(217, 218).

levels. Frequently, however, steel parts are used at high hardness levels, and it is desirable to know what variables govern the mechanical properties at these high hardnesses, and to have a reproducible mechanical test that may be used to insure that the properties have been obtained. By means of simple bend tests of specimens supported at the two ends and loaded in the center, two sources of stress concentration, the fillets at the shoulders of the tensile specimens and the non-axial loading, may be eliminated. Two point loading can be used, if desired, to eliminate artificial predetermination of the position of fracture.

In the bend test the yield strength (obtained from the deviation of the load-deflection curve from linearity) and the bend strength usually are measured. The bend strength is defined as the stress in the outermost fibers (at which fracture of the bar takes place), calculated on the basis that the bar (even after plastic deformation) is completely elastic (i.e., that stress is proportional to strain). The yield and bend strengths for single point loading in the center of the specimen may be obtained by substituting the yielding or breaking loads in the following equations:

$$\text{Strength} = \frac{3\,l}{2a^3}\,L, \text{ square bar}$$

$$\text{Strength} = \frac{8\,l}{\pi d^3}\,L, \text{ round bar}$$

where l is the distance between supports, a the width of the square bar, d the diameter of the round bar, and L the yielding or breaking load.

Before plastic deformation of the bend specimen takes place, the stress varies linearly across the bar, being zero at the center and a maximum at the surface. A slight amount of plastic deformation at the surface does not appreciably raise the stress in the outer fibers, but it does allow the stress to increase within the bar. Thus, the specimen can withstand higher and higher loads as the plastic deformation invades the specimen, for the high stress appears over a greater cross-section of the specimen. The more load the specimen withstands before fracture the higher is the bend strength. As the hardness increases and the yield strength becomes equal to the fracture strength, the bend strength will sharply fall to the yield strength.

The results of typical bend tests are illustrated in Figure 62, in which the bend strength and the yield strength are plotted as functions of hardness. Often, it is desired to obtain the maximum bend strength at the highest possible hardness. This requires the maximum separation between the yield and fracture strengths. At present it appears, that the maximum bend strength is obtained with the "cleanest" tempered martensitic steel of the lowest carbon content necessary to produce the desired hardness(219). No systematic investigation has been made of the relations between yield and fracture strengths at high hardnesses, including the effects of such variables as small amounts of non-martensitic transformation products.

Effects of Temperature and Rate of Strain

The effects of temperature and strain rate on the flow and fracture curves themselves have already been discussed in detail. It is, however, necessary to make some comments regarding the variation of the uniform elongation, tensile strength, and reduction of area with these variables. The effects of the two variables are simple to discuss because the effect of increasing strain rate is equivalent, at least qualitatively, to the effect of decreasing temperature. Thus, only the effect of temperature need be treated. The qualitative effect of strain rate may be derived directly from the discussion of the effect of temperature, as indicated in the preceding chapter.

As the temperature is decreased the strain-hardening exponent increases. This increase in strain-hardening results in a slight decrease of the yield-tensile ratio as the temperature decreases. If nothing occurred to prevent it, it would be expected that the yield-tensile ratio would increase continuously and the strain to necking decrease as the temperature is raised, for the strain-hardening should decrease with increasing temperature. Figure 63 illustrates schematically the effect of temperature upon the load-elongation curves of steels for which the strain-hardening decreases continuously with increasing temperature. These curves are discussed in more detail in the section dealing with creep. It has already been pointed out, however, that in most steels strain aging occurs simultaneously with the deformation at temperatures slightly above room temperature. This progression of aging, deformation, and aging increases the strain-hardening and causes a decrease of the yield-tensile ratio with increasing temperature in the vicinity of 500° to 800° F. (250° to 450° C.) depending upon the rate of straining. Above this temperature, the strain-hardening again decreases and the yield-tensile ratio increases. As illustrated in Figure 64, the tensile strength decreases, increases, and then decreases again as the temperature is raised. The "tensile-strength hump" occurs at higher temperature the higher the rate of strain (Figure 49), presumably because less time is available for precipitation. For sufficiently slow strain rates the increase in tensile strength will occur in the vicinity of room temperature, as discussed in Chapter III.

As illustrated in Figure 64, the yield strength decreases continuously with increasing temperature or decreasing strain rate. In the region of temperature in which the increase in tensile strength occurs, the reduction of area decreases because of the increase in the slope of the stress-strain curve. Because of this decrease in ductility (and at slightly higher temperatures, in notched-bar energy) the phenomenon is referred to as "blue-brittleness"*. None of these effects depends to any great extent upon the gross microstructure of the steel, for blue-brittleness is evident in tests of tempered martensitic steels, as well as of pearlitic.

The variation of the ductility or fracture stress above room tem-

*The oxide coating that forms on steels exposed to air while in this temperature range is blue.

perature has not received adequate study. It can be said, however, that except in a few special cases(220), the reduction of area increases continuously with increasing temperature above the range in which blue-brittleness occurs. The effects of changes in structure on the dependence upon temperature of the fracture stress and reduction of area at high temperatures have received no attention.

The behavior of steels at temperatures below about 70° F. (20° C.) is very dependent upon the microstructure. As the temperature decreases, the flow curve rises faster than does the fracture curve, and the strain-hardening increases. If the fracture stress at small strains is much greater than the flow stress, the strain to fracture (or reduction of area) decreases only gradually as the temperature decreases. If, on the other hand, the fracture stress at small strains is only slightly greater than the flow stress, decreasing the temperature will induce brittle failure. Because of the small separation between the flow and fracture curves at small strains, pearlitic or temper brittle steels break brittlely in the tensile test at low temperatures (about -275° F. (-170° C. or 100° K.)), while tempered martensitic steels of moderate yield strengths do not break brittlely even at extremely low temperatures (-425° F. (-255° C. or 20° K.)). The variation of the reduction of area with temperature below 70° F. (20° C.) is illustrated in Figure 65, for both pearlitic and tempered martensitic structures in a typical steel. A great deal of information exists that confirms this major difference in the mechanical properties at low temperatures of steels of different microstructure.

Notched-Bar Tests

The tensile test does not reveal the major differences between the properties of steels of various microstructures except at rather low temperatures of testing. However, the notched-bar or notched-bar impact test is more sensitive in differentiating between steels of different structures. This test offers the simplest means of comparing experimentally the slopes of the fracture stress curves. Even so, the notched-bar test has been criticized publicly and privately to such an extent that many metallurgists tend to place little reliance upon its results. The results of notched-bar tests, however, reflect differences in behavior that can be quickly and easily interpreted and that can be found by other tests only after numerous and complex experiments.

Description of Tests

Many variations of the notched-bar test are used in different parts of the world, but the Izod and Charpy types have found the most general acceptance in this country. The specimen used in the Charpy test is a square bar notched at the midlength of one face. The specimen is supported by its ends on an anvil and is broken by a pendulum that strikes the face opposite the notch at the midlength of the specimen. The Izod specimen is notched near one end, and broken by a blow striking the notched end. A "V" notch of .01 inch radius is generally used with the Izod specimen, and a "keyhole" type notch with the Charpy specimen. The standard specimens, as described

by the American Society for Testing Materials are illustrated in Figure 66. Many variations of these specimens having notches of various radii, depth, and shape, are used. The round Izod specimen, which is a round bar notched circumferentially, has found rather wide use in testing laboratories in this country. The Charpy specimen with a V notch is also meeting with some approval. Probably the oldest type of notched-bar test is the so-called "nick-break" test, in which a specimen of rectangular cross-section is notched on one or more faces and broken by a blow. Since in this test little care is taken to make the specimens the same size, the energy required for fracture is seldom measured. The type of fracture is used a criterion of the comparative flow and fracture properties of steel. Many years ago the significance of these fractures and how changes in heat treatment cause major differences in fracture types were discussed(222, 223).

The distribution of stress in all such specimens is of the same type. For example, the distribution of the base of the notch in a Charpy bar is essentially the same as in a circular bar notched completely around its circumference and pulled in tension. The distribution of stresses in the latter case has been discussed in some detail(224). A schematic diagram of the stress distribution at the base of the notch in such a specimen is presented as Figure 67. At the very base of the notch, the longitudinal tensile stress σ_x is a maximum. The ratio of this stress at the base of the notch to the value that it would be in the unnotched section is referred to as the stress-concentration factor*.

In Figure 68, the effect of changing the radius and depth of the notch on the stress-concentration factor, b, for a V-notched bar is illustrated. The data for this figure were obtained(225) by photoelastic methods. As the depth of the notch increases or the radius decreases, the stress-concentration factor increases.

Due to this stress-concentration, the metal at the base of the notch tends to contract transversely to a greater extent than the material of the unnotched shoulder; but if the bar is sufficiently wide, the less stressed shoulder essentially prevents the lateral contraction at the apex of the notch. This constraint induces transverse** and radial stresses. The radial stress (σ_y of Figure 67) is necessarily zero at the free surface at the base of the notch. Unless the stress-concentration is small, the magnitude of the transverse stress at the base of the notch depends primarily upon the relation between the width of the bar and the radius of the notch. The wider the bar relative to the radius of the notch, the greater will be the transverse stress. If the bar were infinitely wide, the transverse stress would be sufficient to prevent almost all lateral contraction and in this case the transverse stress would be a maximum.

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*Frequently, the stress-concentration factor has been defined as the ratio of the maximum stress to the average stress across the notched section.

**This transverse stress in a square bar must necessarily be zero at the free surface at the sides of the specimens.

Figure 69 shows the calculated(226) ratio of the transverse tensile stress to the maximum possible transverse stress as a function of the ratio of the radius of the notch to the width of the bar, for a V notch Charpy type specimen. For ratios of radius of notch to width of bar of less than about 1/10, the transverse stress is essentially at its maximum and changes in the bar width will have little effect on the transverse constraint.

At the base of the notch of the V-notched Charpy bar the longitudinal stress required for flow, as determined by the stress distribution, has been calculated(191) to be:

$$\frac{\sigma'_x}{\sigma'_{x(c)}} = 1 - \frac{\gamma(b-1)}{b} - \frac{[\gamma(b-1)]^2}{b} \qquad (4\text{-}10)$$

where σ'_x is the tensile stress required for flow in simple tension, $\sigma'_{x(c)}$ for yielding in the presence of the transverse stress, and γ is Poisson's ratio for plastic deformation (.5). The relation between the increase in yield strength and stress-concentration factor derived from this equation is illustrated in Figure 70. For the case of the V-notched bar with .01 inch radius the figure indicates the effect of the notch is to raise the flow curve by about 11% (stress-concentration factor approximately equal to 6). The general principle that the transverse stress induced by the notch raises the longitudinal stress necessary for flow applies to any type specimen.

The gradient in stress beneath the notch may also raise the stress required for plastic flow(161). The sharp stress gradient (Figure 67) may be the prime cause of the size effect in notched-bar tests so often suggested*. The stress gradient at the base of the notch is different in similar specimens of different sizes. The precise effect of this gradient cannot be taken into account, however.

The transverse stress induced by the notch raises the longitudinal tensile stress necessary for plastic flow. However, the fracture-stress curve does not appear to be affected by the presence of the transverse stress. As discussed in Chapter III, the available data indicate that fracture occurs, after a given amount of deformation, when the tensile stress reaches a fixed value. The effect of the notch, therefore, is to raise the flow curve with respect to the fracture curve.

Because of the stress-concentration at the base of the notch, the strain rate is higher for a given rate of loading than would be found with an unnotched specimen. The strain rate at the base of the notch is equal to the stress-concentration factor multiplied by the strain rate of the material in the unnotched shoulder. Changes in notch radius or notch depth must be relatively large to induce significant changes in stress concentration (Figure 68) and hence in strain rate.

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*An apparent size effect can arise because of differences in metallurgical structure due to differences in transformations between specimens heat-treated in different sizes.

The effect of the increased strain rate is to raise both the flow and the fracture curves, affecting the fracture curve less than the flow curve. The difference in strain rate between that of the standard tensile and the notched-bar-impact test is tremendous and its effect is significant. The strain rate at the base of the V notch in a Charpy bar has been calculated(191) to be about 10^3 sec.$^{-1}$ (approximately 10^7 times that encountered in the standard tensile test).

Significance of Results

The notched-bar test differs from the normal tensile test in that the metal at the base of the notch is constrained from transverse contraction and the strain rate is many times greater. The effects of these factors on the flow and fracture stress curves of steels of typical structures is illustrated schematically in Figure 71. The increased strain rate over that normally encountered in tensile tests raises the flow curve more than it does the fracture curve. The transverse stress raises the flow curve but does not affect the fracture curve. If the separation between the flow and fracture stresses at small strains is large, the raising of the flow curve with respect to the fracture curve will induce only a small decrease in strain to fracture. On the other hand, if the separation between the flow and fracture stresses is small, the strain to fracture may decrease markedly (the amount depending upon the shape of the fracture stress curve). If the separation between the flow and fracture curves (at small strains) is small enough, the effects of the transverse stress and the high strain rate may be sufficient to induce brittle failure at room temperature (or even above). Decreasing the temperature will further raise the flow curves as illustrated in Figures 72 and 73. If the separation between the flow and fracture curves is large, the strain to fracture will decrease only gradually as the temperature is lowered. If the separation between the flow and fracture curves is small (and the dependence of the fracture stress on strain is marked), the strain to fracture will undergo a rather sudden decrease (to very small values) as the temperature is decreased.

From the schematic stress-strain diagrams of Figures 70 and 71, the variation of the energy absorption of steel specimens with temperature may be estimated. Schematic curves relating the impact energy with temperature of test are also presented in Figures 72 and 73. As the temperature is lowered for steels having a small separation (at small strains), between the flow and fracture stresses the energy suddenly decreases as brittle failure occurs over at least part of the specimen. At still lower temperatures the entire specimen breaks brittlely (very small energy required for fracture). The fracture changes from "fibrous" (or ductile), to the picture-frame type, to completely "crystalline" (bright). These fractures are illustrated in Figure 74. As the temperature is lowered for steels having a high and nearly horizontal fracture-stress curve, the energy necessary for fracture decreases relatively gradually. A gradual change of fracture also occurs: from fibrous to dull "crystalline", with an intermediate type of picture-frame fracture having a dull crystalline center surrounded by a fibrous border. The change of fracture from fibrous to dull crystalline occurs at much lower temperatures than the change from fibrous to bright crystalline (for steels of the same tensile strength).

The area under the stress-strain curve represents the energy required for fracture per unit volume of material at the base of the notch. The larger the amount of the deformation of the metal at the base of the notch, the greater the volume of the specimen that is plastically deformed. Thus, as the strain to fracture decreases, the amount of deformed metal at the base of the notch also decreases. The energy per unit volume decreases and so does the volume of metal affected. The total energy absorbed thus decreases sharply.

The analysis of the behavior of steels in the notched-bar test has to this point considered only the initiation of fracture. As the crack caused by the fracture proceeds across the specimen it becomes not only deeper but sharper; the stress-concentration factor and the constraint increase and the flow curve rises. Thus, even if brittle failure does not occur at the base of the performed notch, it may occur as the fracture proceeds. Furthermore, since at the sides of the specimens the transverse stress is necessarily zero (there is no constraint), the conditions leading to brittle failure are not as severe as at the center of the bar. "Picture-frame" fractures (arising from these causes), as illustrated in Figure 74 are frequently observed with pearlitic and bainitic steels*.

As indicated in Chapter III, the fracture stress at small strains is related primarily to the type of structure. Thus, the notched-bar properties are markedly different for steels with the same strengths but different structures. Pearlitic steels in general break brittlely at or near room temperature while tempered martensitic steels have been observed to maintain fair toughness(184) down to temperatures as low as -425° F. (-255° C. or 20° K.). Steels having bainitic or tempered bainitic structures generally have notched-bar properties intermediate between those of tempered martensite and pearlite; the dependence of the fracture stress on strain is, therefore, intermediate between those of pearlitic and of tempered martensitic steels.

Thus, at a given hardness, steels having tempered martensitic structures have superior properties as measured by the notched-bar test. The advantage of this structure can arise only from a difference in height and shape of the fracture stress curve, which appears to be connected with the different shape, size, and distribution of the carbide particles.

Even though our primary interest is in the properties of tempered martensitic steels, it is desirable to discuss the effect of variables on the notched-bar properties of pearlitic steels. The finer the pearlite plates, the higher is the fracture stress for a given yield strength and hence the lower is the transition temperature. As the hardness of pearlitic steels is reduced by tempering, the separation between yield and fracture strength increases, the slope of the fracture strength curve decreases; the

*The edge away from the notch has a fibrous or ductile appearing fracture probably because it suffers deformation in compression while the metal at the base of the notch is being plastically deformed in tension.

notched-bar properties improve. In fact, the transition temperature can be decreased to such low temperatures that only very severe tests will distinguish between the properties of pearlitic and tempered martensitic steels. It is well known that with low-carbon air-cooled steels, the addition of nickel improves the notched-bar properties by decreasing the temperature of brittle failure(227). It is possible that this is a specific effect of nickel, but probably nickel (like other alloying elements) by slowing the pearlite transformation causes finer carbide particles to be produced for a given air-cooling treatment. Cold working of pearlitic steels is an effective way of reducing the separation between the flow and fracture stresses and of thus raising the temperature of brittle failure(228)*.

The properties of steels having bainitic structures are difficult to establish, for frequently the retention of austenite accompanies the formation of bainite. Subsequent tempering may cause this austenite to transform to high temperature transformation products which lower the notched-bar properties. Very probably, the lower the temperature of formation of the bainite, the better are its notched-bar properties after tempering. The structure resulting from the high temperature tempering of bainite is practically indistinguishable from that of tempered martensite and the mechanical properties of the two structures are much the same. Since, however, the bainite has a distinct orientation of the carbide particles, its notched-bar properties after lower temperature tempering treatments are inferior to those of tempered martensite of the same hardness and carbon content. As little as 10 per cent of tempered bainite (formed at high temperature) in an otherwise tempered martensitic steel will noticeably decrease the notched-bar energy at -40° F.**.

The notched-bar curves of Figure 75 indicate the effect of yield strength upon the notched-bar properties of a typical tempered martensitic steel. As the carbides become finer and more numerous the level of the notched-bar energy decreases and the temperature of the decrease in notched-bar energy rises slightly. Since the height of the fracture stress curve depends upon carbon content (for a fixed yield strength) the notched-bar energy decreases as the carbon content increases, at a constant yield strength. However, the notched-bar properties are practically independent of alloy content if the steels have completely martensitic structures on quenching and if subsequent heating operations do not induce the precipitation that is the cause of temper brittleness. The variations from heat

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*There is a report(229) that cold working by drawing lowers the temperature of brittle failure of unnotched bars while the temperature is lowered for notched bars. It is probable that the decrease in the temperature of brittle failure was associated with the introduction of surface compressional stresses by the cold deformation. The specimens were not machined before testing.

**Based upon unpublished data obtained at the Watertown Arsenal Laboratory. These tests were performed with .30% carbon steels at 260 Brinell hardness. The bainite was formed at relatively high temperatures upon continuous cooling.

to heat are of the same order of magnitude as the variations among steels of different alloy contents. Data showing the variation of notched-bar energy (at room temperature) are presented as Figure 76. The specimens for these tests were taken from bar stock (parallel to the axis of the bars) of steels of carbon contents ranging from about .25 to .45%. Notched-bar data as a function of hardness for a large number of specimens taken parallel and perpendicular to the principal rolling direction of cross-rolled plate are included in Figures 77 and 78. More scatter is evident in these data than in those relating the reduction of area with the tensile strength. Since the notched-bar energy is a very sensitive measure of the height and slope of the fracture stress curves, small variation of these curves are reflected in major variations of notched-bar energy. At moderately high strength levels small differences in the level of the fracture stress cause great differences in the energy required to break notched specimens, just as (at higher hardnesses) small differences reflect themselves in large changes in bend test results. At high strength levels a small lowering of the fracture stress curve may result in completely brittle failure and the notched-bar test is unsatisfactory for determining the properties of steel under these conditions. Slow bend tests of unnotched specimens are much more reproducible and hence more satisfactory.

The discussion in the foregoing paragraphs is concerned with the properties of tempered martensitic steels. With such steels, as the carbides become larger as a result of tempering, the fracture stress decreases less rapidly than does the flow stress, and the notched-bar energy increases. (Figures 75 to 78.) In the transition from untempered martensite to tempered martensite, the fracture stress may decrease more rapidly than the flow stress and may cause an initial decrease in notched-bar energy as the steel softens (Chapter VIII).

Differences in the slope or level of the fracture stress curves arising from inclusions also affect the notched-bar properties. Differences between transverse and longitudinal specimens of tempered martensitic steels are illustrated in Figures 77 and 78. Some inclusions appear to have specific effects; aluminum nitride, for example, which appears to concentrate at austenitic grain boundaries, induces brittle failure at relatively high temperatures(230). A conchoidal type of fracture is associated with these low energy failures. Steels containing high silicon contents have also been reported to have low impact properties, presumably because of the formation of silicates(30). Defects such as flakes, voids, or large inclusions may cause stress-concentrations that may initiate fracture. The notch effect of such defects may be more severe than that of the mechanical notch. In service, these defects may be the only notches present.

Besides the direct effect of inclusions and voids on the fracture stress, they may frequently cause laminations that markedly change the conditions of the notched-bar tests. During the rolling of plate, for example, inclusions are rolled out in planes parallel to the plate. The tensile stress required for fracture on these planes may be low and in the notched-bar test they open. If the laminations are parallel to the long axis of the specimen and perpendicular to the notch, opening of a lamination by

the transverse stress will effectively cause the formation of two specimens, both of which are narrower and less constrained than the original specimen. Brittle failure will then occur only at lower temperatures than for an unlaminated specimen. Laminations parallel to the notched surface will often cause the crack to stop, and a new crack to begin, reducing the overall severity of the test. The temperature of brittle failure should be the same but the transition zone from ductile to brittle failure extends over a wider range. If the laminations are perpendicular to the length of the bar, fracture may take place along a lamination and markedly decrease the energy absorption.

It may be deduced from the results of impact tests that increases in prior austenitic grain size lower the fracture stress curve and perhaps increase its slope. These changes are reflected in a decrease in notched-bar energy at high testing temperatures(231) and an increase in the temperature of transition from ductile to brittle failure. It is controversial, however, whether this effect is directly connected with the increase in grain size or whether it arises from a change in size, shape, or distribution of carbides induced by the increase in grain size of the austenite.

The temper brittleness precipitate also has deleterious effects upon the notched-bar properties. The presence of the precipitate lowers the fracture-stress curve as illustrated in Figure 34. Embrittlement by this precipitate is caused by a change in slope of the fracture stress-strain curve and is not ordinarily reflected in tensile tests at or near room temperature. While the precipitate raises the temperature of transition from ductile to brittle failure, its effect upon the notched-bar energy depends upon the temperature of test and upon the properties of the unembrittled steels. Thus, the effect of a fixed amount of precipitate on the notched-bar energy absorption at high testing temperatures is more apparent in pearlitic steels than in tempered martensitic steels of the same strength (Figure 79).

As has already been discussed, the temper brittleness precipitation occurs at temperatures below 1100° F. (600° C.) at a rate that depends upon the temperature and the composition of the steel. Thus, for steels heated in the embrittling temperature range the notched-bar properties depend upon composition; the higher the manganese, nickel, and chromium contents and the lower the molybdenum content the greater is the embrittlement with a given treatment (for short and moderate times). If, however, tempering is carried out above about 1100° F. (600° C.) and the cooling following tempering is rapid enough to avoid the temper brittleness transformation at the nose of the "C" curve (Figure 18), embrittlement will be avoided. Part of the scatter in the impact values of Figures 76-78 is probably the result of varying degrees of temper brittleness of the various steels. Certain recent data indicate that, if no temper brittleness (or decomposition of retained austenite) occurs, a steel tempered to a given yield strength has the same reduction of area, tensile strength and notched-bar properties whether the tempering is at a high or at a low temperature*.

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*Unpublished work.

The notched-bar properties of steels above room temperature are relatively independent of metallurgical structure. In Figure 80 the notched-bar energy of a typical pearlitic steel as a function of temperature is presented. The impact energy increases up to about 200° F. (100° C.), then decreases, reaches a minimum, then increases again. This decrease in notched-bar energy in the range from 200 to 900° F. (100 to 500° C.) is another manifestation of blue-brittleness. Because of the high strain rates encountered in the notched-bar impact test, the drop in toughness occurs at a higher temperature than does the decrease of ductility in the tension test. Low carbon steels to which special agents have been added do not exhibit blue-brittleness to a major degree(157). The increase in toughness at still higher temperatures is a manifestation of the decrease in strain-hardening arising from recovery and recrystallization.

Fatigue Tests

Not only must metal parts withstand the stresses for which they are designed but frequently they must withstand these stresses for an enormous number of applications of the load. The behavior of actual parts in service would be extremely difficult to predict even if the knowledge of the fatigue properties of metals was more complete than it is, for it is almost impossible to estimate precisely what the service conditions will be. Since parts are not subjected to simple tensile loads nor are they usually machined with the care of laboratory test specimens, the number of load applications which the part will withstand is difficult to assay. Generally, all that can be done is to evaluate the relative behavior of different metals under conditions of the same kind as those met in the service.

Frequently the possibility of fatigue fracture may be eliminated by a change in mechanical design such as the elimination of notches or sharp fillets. Increasing the fatigue life of metal parts by introducing favorable stresses through "shot-peening" has recently been widely utilized. Even if all practical mechanical steps have been taken to reduce the danger of fatigue fractures, it is frequently desired to increase the life of the part by improving the fatigue properties of the metal. In order to do so it is necessary to understand the general principles governing the effects of repeated cyclic deformation on steel and the relation between metallurgical structure and the response of steel to such deformation. The fatigue properties of steel have been studied in metallurgical laboratories for many years but little systematic knowledge is available relating the fatigue properties to the stress-strain curves or to metallurgical structure. Many fatigue data are available for specific steels, however.

Description of Tests*

*The most generally used tests are described in the book "The Fatigue of Metals"(232).

Fatigue tests are usually performed in bending, tension and compression, or torsion. Generally, they are performed by applying a complete reversed-stress cycle. For example, the specimen is loaded in tension, unloaded, loaded in compression, unloaded, and the cycle repeated. The entire cycle is continuous and generally repeated thousands of times per minute. Bend and torsion tests are simpler to perform than tension-compression tests and as a consequence have been more frequently used. In one of the oldest types of test a round bar is rotated by a chuck fastened at one end while the other end is loaded by a spring through a bearing sleeve. The stress at the surface of the specimen varies sinusoidally with time from tension to compression. The amplitude of the stress is determined by the constant tension in the spring, the frequency by the speed of the rotation. Many modifications of this rotating-beam machine have been used throughout the world. With all such constant-load machines the number of cycles required for failure are determined for a number of values of the load. Curves of the type illustrated in Figure 81 are obtained for steels and are generally called simply "S-N" curves, where S is the stress applied and N the number of cycles. When plotted on logarithmic paper as in Figure 81, the relation between S and N is approximately linear. Below a certain definite value of the stress, however, an infinite number of cycles may be applied to steel without causing failure. This value of the stress is referred to as the endurance limit*. The endurance limit frequently must be considered in the design of steel parts and fortunately it bears a rather simple approximate relation to the tensile strength. S-N curves are affected by changes in the surface finish of the specimens and by the presence of notches and fillets. In any attempt to find the effect of metallurgical variables these conditions of the test must be held as nearly constant as possible.

Significance of Results

A complete understanding of the mechanism of fatigue failure must await an atomistic interpretation of strain-hardening and fracture. The ordinarily measured stress-strain curves indicate that below a certain stress value plastic flow does not occur to an appreciable extent. Further, if a stress is applied that causes plastic flow, a second application of the same stress should not cause further flow or further strain-hardening. Below a certain limiting stress, therefore, plastic flow should not occur, and further, according to the stress-strain curve, fracture should not occur at stresses below that at which the flow and fracture stress curves intersect. However, upon repititive stressing, failure occurs much below the ordinarily measured yield stress (even if determined at small plastic strains) and at stresses much below the fracture stress as determined in simple tension. Thus we must look to microscopic inhomogeneities (stress concentrations) for the mechanism of failure. As has been pointed out, "a bar of uniform strength, whose surface was perfectly smooth with no sharp corners in the longitudinal configuration, and with a perfectly

*All metals do not have definite endurance limits; the number of cycles increases continuously as the stress is lowered for many non-ferrous metals(232).

homogeneous structure, would endure without breaking an indefinite number of repetitions of the stress varying between zero and a value near the yield strength"(233).

Below the endurance limit some local flow occurs initially but does not appear to continue for a large number of cycles(234, 235). During fatigue it appears that residual stresses on a microscopic scale are introduced if the stress is sufficiently high relative to the yield strength(177). Additional applications of stress cause plastic flow on a local scale which increases the residual stresses until failure occurs. If the applied stress is not sufficiently high, microstresses do not appear to be introduced (or, in fact, may be reduced). Whether or not residual stresses are introduced appears to depend upon both the yield strength and the rate of strain-hardening. Thus, correlations of the endurance limit directly with the yield strength lead to certain inconsistencies. It is found that the endurance limit bears a definite relation to the tensile strength (Figure 82), rather than to the yield strength as ordinarily determined, for the tensile strength also increases as the strain-hardening increases. Another reason for the lack of correlation with the yield strength is that in repetitive loading steel behaves as if it yielded homogeneously even though it exhibits a drop of load at yielding in the tension test(236). Since the tensile strength is not affected by the inhomogeneous yielding, it, again, is more directly related to the endurance limit than is the yield strength. The best and simplest expression of the relation between endurance limit of steels, $\sigma\varepsilon$, and tensile strength S_0, is that:

$$\sigma\varepsilon = \frac{S_0}{2} .$$

Since the endurance limit is dependent only upon the tensile strength, the endurance limits of steels of various structures are nearly the same if their tensile strengths are the same. As the strength of steel increases as a result of reducing the tempering temperature to low values, the endurance limit rises less rapidly with tensile strength as is indicated by Figure 82. For steels tempered at very low tempering temperatures, the endurance limit may in fact be lower than that of softer steels tempered at higher temperatures. Tempering at low temperatures may not reduce the quenching stresses (Chapter VIII) and the applied stress necessary to produce a slight amount of plastic flow may then be very low.

The endurance limit may also be decreased by a small prior plastic deformation. Loading the metal repetitively below the endurance limit (called "understressing") for a large number of cycles will restore or even raise the endurance limit, lowered by a small amount of prior deformation(178).

If the alternating stress is not too large (compared to the average yield strength), multiple applications of the stress cause the reduction of stress peaks by plastic flow in restricted regions. This process continues until the metal becomes almost completely elastic (linear stress-strain curve with little hysteresis). The readjustment of the specimen in the local regions where deformation has occurred is not sufficient to cause uniform plastic deformation. As is pointed out in the discussion on the

relief of the internal stresses (Chapter VIII), the yield strength of deformed metals may be increased by the removal of stress peaks by understressing. However, if the stress is sufficiently high, the residual stresses will be increased.

If instead of stressing specimens below the endurance limit, a number of cycles of high stress insufficient for failure are applied ("overstressing"), it is found that the endurance limit is reduced*. So-called "damage" lines have been determined that indicate the number of cycles at various stress levels required to do a specified amount of damage to the endurance limit. Figure 83 indicates the number of cycles at various stress levels necessary to produce fixed percentage reductions of the endurance limit. If the damage is not too severe subsequent understressing may raise the endurance limit to or above its original value.

If steel is repeatedly stressed above its endurance limit, local plastic flow occurs on each cycle, and local stresses are introduced until the fracture stress is reached locally. The crack thus formed progresses across the specimen. Thus the number of cycles to failure above the endurance limit must depend upon the difference between the flow and fracture stresses at small strains, and therefore upon metallurgical structure. The number of cycles to failure should be related to the results of notched-bar impact tests. However, in most fatigue tests the frequency of loading is so great that the temperature is raised appreciably above the ambient temperature. The deformation is essentially adiabatic. The conditions of the test may not be severe enough to distinguish between steels of different structures unless their properties are widely different. At a given tensile strength, tempered martensitic steels should endure the maximum number of cycles to failure above the endurance limit in ordinary fatigue tests. As the structure is modified so that the fracture stress at small strains decreases and the temperature of brittle failure in the notched-bar impact test increases, the number of cycles to failure in fatigue decreases. As an example, the effect of the temper brittleness precipitate on the S-N curves is illustrated in Figure 84. The notched-bar impact properties for the same steels are also plotted in the figure. The higher the transition temperature in the notched-bar test, the smaller the number of cycles above the endurance limit required for fatigue fracture. The transition temperature is raised by the lowering of the fracture stress curves (Figure 34). It is to be noted that for the steels of Figure 84, the endurance limit, which depends only upon flow characteristics, is nearly unaffected by the temper brittleness precipitate, for the precipitate does not affect the flow-stress curves.

Thus, all these factors that affect the notched-bar impact properties affect the number of cycles to failure in fatigue**. As the yield

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*An excellent discussion of the effects of overstressing is to be found in a recent book(239).

**The effects of these factors may be small as reflected in the results of the ordinary high frequency tests.

strength is increased the entire S-N curve rises. At a given stress value the number of cycles to failure increases. The effect of changing the strength level of tempered martensitic steels is illustrated in Figure 85. Also in this figure the logarithm of the ratio of the applied stress to the flow stress at a small strain (from tensile data) is plotted as a function of the logarithm of the number of cycles to failure. This method of plotting makes for easy comparison of the data for the different strength levels.

It should also be noted that failure by fatigue usually takes place in two distinct stages: the formation of the fatigue crack, and the progress of this crack across the specimen. At low values of the stress (large number of cycles) the number of cycles required to form the crack are relatively large compared to the cycles required to make it transverse the specimen. At high stress levels the number of cycles required to start the crack may be of the same order as the number required for the crack to progress across the specimen. The progress of the crack across the specimen is governed by the same factors that govern the behavior in notched-bar tests. As the crack becomes sharper and deeper the stress concentration increases, and the flow curve is raised with respect to the fracture curve. The crack in a steel having a high transition temperature in the notched-bar test may become deep or sharp enough for brittle failure of a large part of the section. The fracture consists of two parts: "1. a smooth surface where the crack has spread slowly and the walls of the crack are battered smooth by repeated opening and closing, and 2. a rough 'crystalline' surface indicating the very much more sudden fracture of the core of the piece"(241).

The fatigue crack in tempered martensitic steels, on the other hand, may continue to grow by very small increments without ever causing sudden brittle fracture. As the section is reduced by the progress of the crack and the load kept constant, the average stress increases and the depth of the crack increases by larger and larger increments until the remaining section finally fractures ductily (fibrously). Thus, at very large loads, where the number of cycles necessary to initiate fracture is small compared to the cycles necessary to spread it, the number of cycles to failure depends even more markedly on the metallurgical structure than at smaller loads. Tests illustrating this relation have been carried out recently(242).

Fatigue fractures usually start at the surface of metal parts and the homogeneity, strength, and contour of the surface markedly affect the initiation of fatigue failures. The presence of surface defects or stress concentrations such as scratches and notches lowers the entire S-N curve, since it is at such stress concentrations that failure begins. The effects of notches, fillets, and surface preparation have been studied and empirically established(232, 239). In general, the effect of notches in reducing the endurance limit is less than would be predicted from calculations based on elastic deformation. Some plastic deformation does occur which reduces the stress concentration. For practical purposes, however, stress concentrations are to be avoided as much as possible in the design of steel parts to be subjected either to impact or to repeated loads.

The endurance limit of steel parts or specimens may be improved by introducing "negative stress concentrations" - surface compressional stresses - since fracture can only take place in tension. By this means, the tensile

stress at the surface may be reduced for any given operating load. Residual compressive stresses may be introduced by peening and by other methods. The performance of steel parts used in reciprocating machinery has been improved by shot peening(243).

Effect of Complex Stresses

Since for ductile metals, at least, the endurance limit is related directly to the stress required to produce a small amount of plastic strain, it might be expected that the relation between the stresses necessary for failure at the endurance limit would be given by the strain energy relation (equation (3-7)). It has been found(244,245) that for metals ductile in the ordinary tensile test, this relation between the stresses at the endurance limit fits the data better than any other. Moore has pointed out, however, that for normally brittle metals, such as cast iron, at the endurance limit failure occurs, under complex stresses, when the maximum tensile stress reaches a critical value. In metals, such as cast iron, in which practically no plastic flow can take place before fracture, it seems that the fracture stress, rather than the flow stress, governs the endurance limit, for as has been pointed out in an earlier section, fracture appears to occur when the maximum tensile stress reaches a critical value.

Few, if any, data are available regarding the effect of complex stresses on regions of the S-N curve other than the endurance limit.

Effects of Frequency of Loading and Temperature

Unless the frequency of loading is changed by a large factor, little change in the endurance limit is produced(246). Increasing the frequency tends to raise the endurance limit(247), just as increasing the rate of strain raises the flow stress curve. At the frequencies generally employed in fatigue tests, the deformation is adiabatic and for stresses much in excess of the endurance limit the temperature of the specimens increases markedly. This increase in temperature should tend to increase the number of cycles to failure and to reduce differences between the number of cycles to failure for steels of different structures.

The effect of temperature on the endurance limit is similar to its effect on the tensile strength. The endurance limit appears to decrease almost continuously as the temperature is raised(248), although the increase in strain-hardening increases it slightly in the vicinity of the "blue-brittle" range(249, 250). At very low temperatures the difference between the number of cycles to failure for steels having different microstructures should be even greater than at room temperature. At very high temperatures steels of different structure should have similar fatigue properties, but so much "creep" occurs under the action of small stresses that fatigue failures are not generally a problem at very high temperatures.

Creep Tests

Steel parts are frequently used in machines that operate at high

temperatures. Sometimes loads are applied at the high temperatures and immediately removed or reversed. Under these conditions the yield strength must not be exceeded, and if the load is applied a large number of times, the endurance limit at the operating temperature may possibly be the important criterion of service life. More often, however, the chief problem at high temperatures is to keep plastic flow to some safe engineering minimum. Even though the yield strength as determined by short time tensile tests is not exceeded, anelastic deformation and plastic flow under sustained loads will occur. Generally speaking, the ordinary heat-treated steel parts will not support appreciable load at extremely high temperatures without appreciable creep. Furthermore, at extremely high temperatures, resistance to corrosion and oxidation becomes a special design problem. For very high temperature service special alloying elements are usually added to induce a high creep resistance at the operating temperatures. Ordinarily heat-treatable steels are not generally used at temperatures much in excess of 1000° F. Since we are concerned primarily with this type of engineering steels and not with those designed specifically for high temperature service, the discussion of creep will be brief*.

A very excellent approach to an understanding of the creep problem has been developed(146). In Figure 86 are plotted the load-elongation curves for a steel at various rates of loading at some elevated temperature. (It will be assumed, for the purposes of the present discussion, that strain-aging - blue-brittleness - does not occur.) At the higher rates of loading the strength is high and the strain at which necking occurs is large because the strain-hardening is large. As the strain rate is decreased the load-elongation curve is lowered and the strain at which necking occurs becomes smaller (the strain-hardening decreases). Now if a number of specimens are tested at various values of the load, their behavior can be predicted qualitatively by constructing horizontal constant-load lines in the top diagram of Figure 86. The strain rates at various values of the elongation may be determined from the intersection of the horizontal line with the load-elongation curves. Creep rate vs. elongation curves constructed in this way for various values of the load are also plotted in Figure 86. This construction assumes that the strength is dependent only upon the instantaneous values of the strain rate and the strain, and is independent of the past history of the deformation. This assumption has not been justified by experiment and cannot be true over wide ranges of strain or strain rate**, but the results of this approach are qualitatively in accord with the observed creep behavior. Elongation-time curves for constant loads may be constructed by integrating the creep rate vs. elongation curves, and are presented in the bottom diagram of Figure 86. It has been pointed out(251) that the series of curves of Figure 86 can apply to different temperatures as well as to different loads. The effect of changing the temperature is qualitatively equivalent to the effect of changing the load.

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*More thorough discussions of the creep problem may be found in several references(251-253, 177).

**The validity of the assumption is discussed in detail in a recent paper(177).

The creep curves (elongation-time curves) of Figure 86 are generally divided into three regions(Figure 87); one in which the deformation occurs at a decreasing rate (primary creep), one in which the rate is approximately constant (secondary), and one in which the rate increases rapidly (ternary). The first stage consists of plastic flow and probably anelastic deformation, the second stage of uniform plastic deformation, and the third stage of localized flow -- "necking" and perhaps local fracture (by grain boundary separation).

The creep rate depends primarily on the strength and the strain-hardening of the metal as a function of temperature and rate of strain. The strain-hardening in turn depends upon the strength of the metal, the rate of recrystallization and recovery, and the presence of precipitation effects (as functions of temperature and strain rate). If any type of precipitation occurs that progressively strengthens the metal, the strain-hardening is increased and the creep rate decreased. In steels, the process of secondary hardening through the formation of complex carbides is probably responsible for the superior creep resistance of steels containing carbide-forming elements. If, on the other hand, transformations occur at the testing or service temperature that reduce the strength, the creep resistance will be decreased. For example, softening arising from the agglomeration of carbides increases the creep rate. Sometimes, under certain conditions at elevated temperatures, chemical reaction of the metal with the surrounding medium occurs and embrittles the metal.

The discussion above is concerned with plastic flow as a result of sustained loads. Frequently, however, the nature of the fracture varies with temperature and magnitude of the applied load. If, for example, at elevated temperatures the precipitation that causes temper brittleness occurs, the fracture-stress curve will be lowered, causing a premature fracture at the elevated temperature, or brittle fracture at stress concentrations upon subsequent loading at low temperatures. At elevated temperatures brittle intercrystalline fractures have also been observed(251) and are frequently of importance.

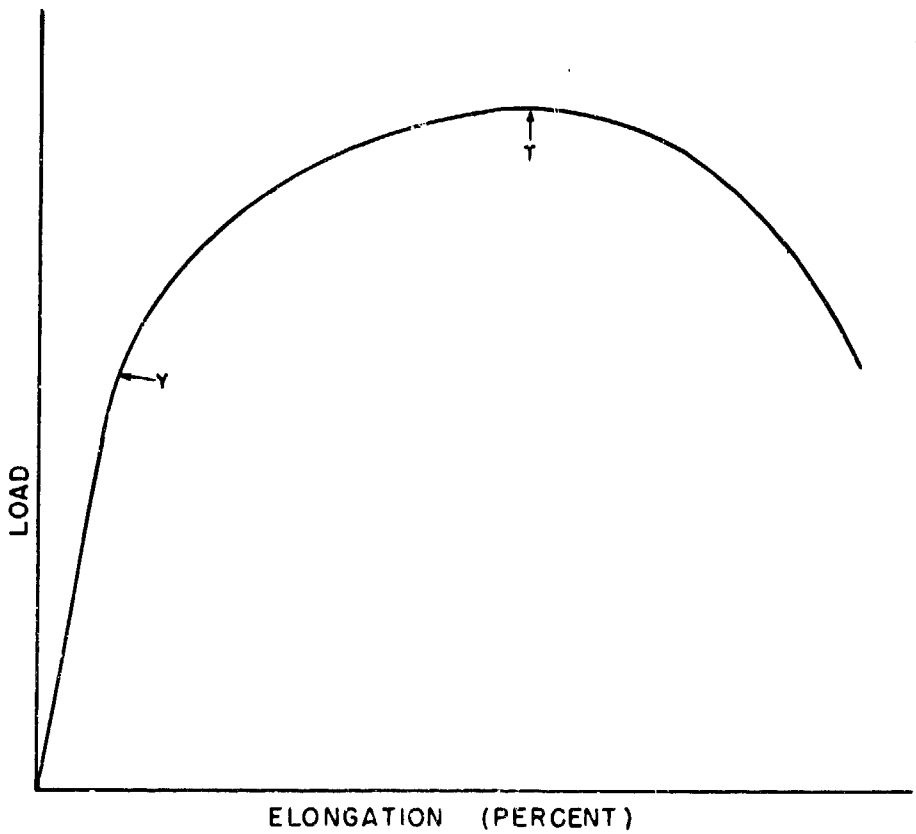


FIGURE 54

SCHEMATIC LOAD-ELONGATION DIAGRAM

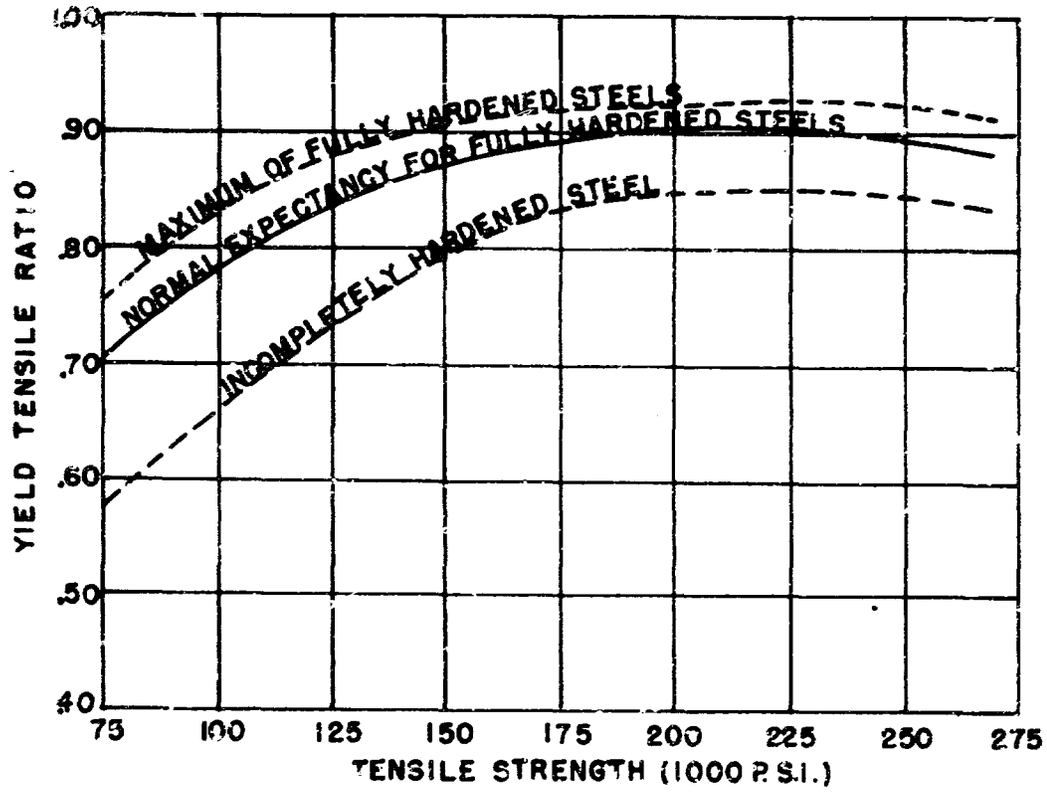


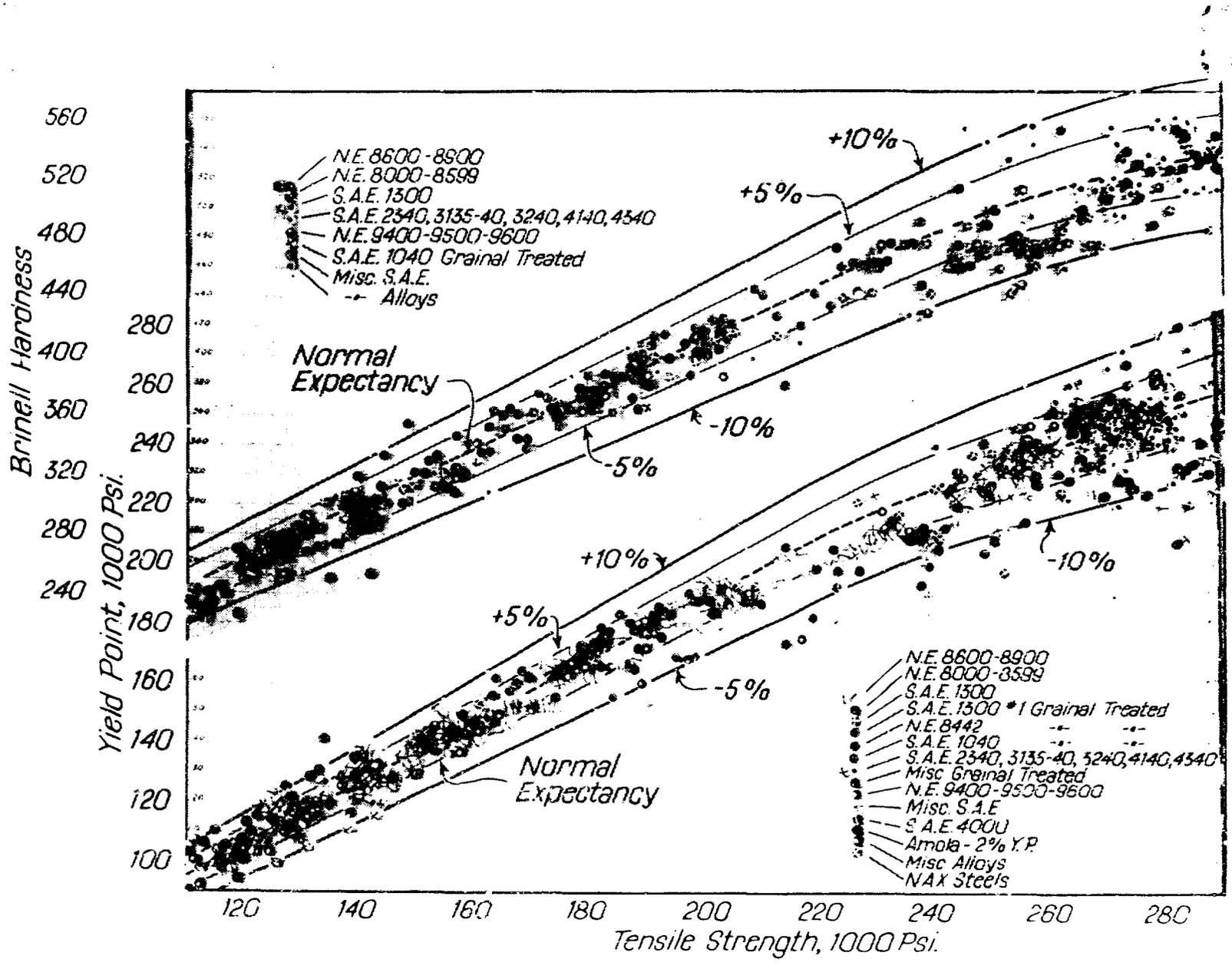
FIGURE 55

VARIATION OF YIELD-TENSILE RATIO FOR STEELS OF DIFFERENT STRUCTURES. (FROM SAE HANDBOOK[207])

FIGURE 56

RELATIONS BETWEEN YIELD STRENGTH AND TENSILE STRENGTH AND BETWEEN BRINELL HARDNESS AND TENSILE STRENGTH FOR TYPICAL COMMERCIAL STEELS (.25 - .45% CARBON)

(AFTER PATTON(208))


Brinell Hardness
560
520
480
440
400
360
320
280
240
Yield Point, 1000 Psi.
280
260
240
220
200
180
160
140
120
100
Tensile Strength, 1000 Psi.
120
140
160
180
200
220
240
260
280
N.E. 8600-8900
N.E. 8000-8599
S.A.E. 1300
S.A.E. 2340, 3135-40, 3240, 4140, 4340
N.E. 9400-9500-9600
S.A.E. 1040 Grainal Treated
Misc. S.A.E.
Alloys
Normal Expectancy
+10%
+5%
-10%
-5%
N.E. 8600-8900
N.E. 8000-8599
S.A.E. 1300
S.A.E. 1300 #1 Grainal Treated
N.E. 8442
S.A.E. 1040
S.A.E. 2340, 3135-40, 3240, 4140, 4340
Misc Grainal Treated
N.E. 9400-9500-9600
Misc. S.A.E
S.A.E. 4000
Amola - 2% Y.P.
Misc Alloys
NAX Steels

WTN.639-7568

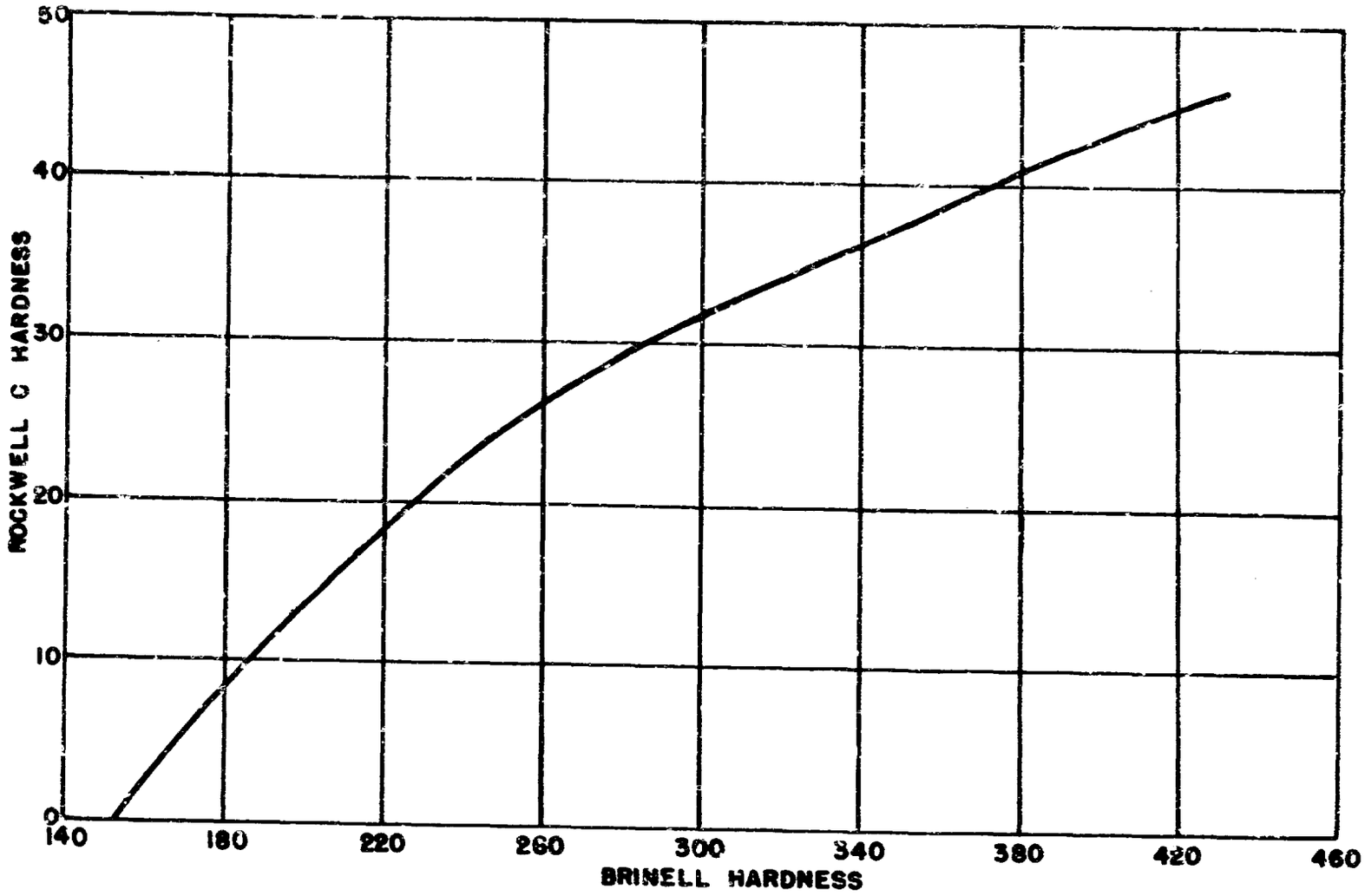


FIGURE 57

RELATION BETWEEN BRINELL & ROCKWELL C HARDNESS NUMBERS
(FROM DATA OF SAE HANDBOOK[207])

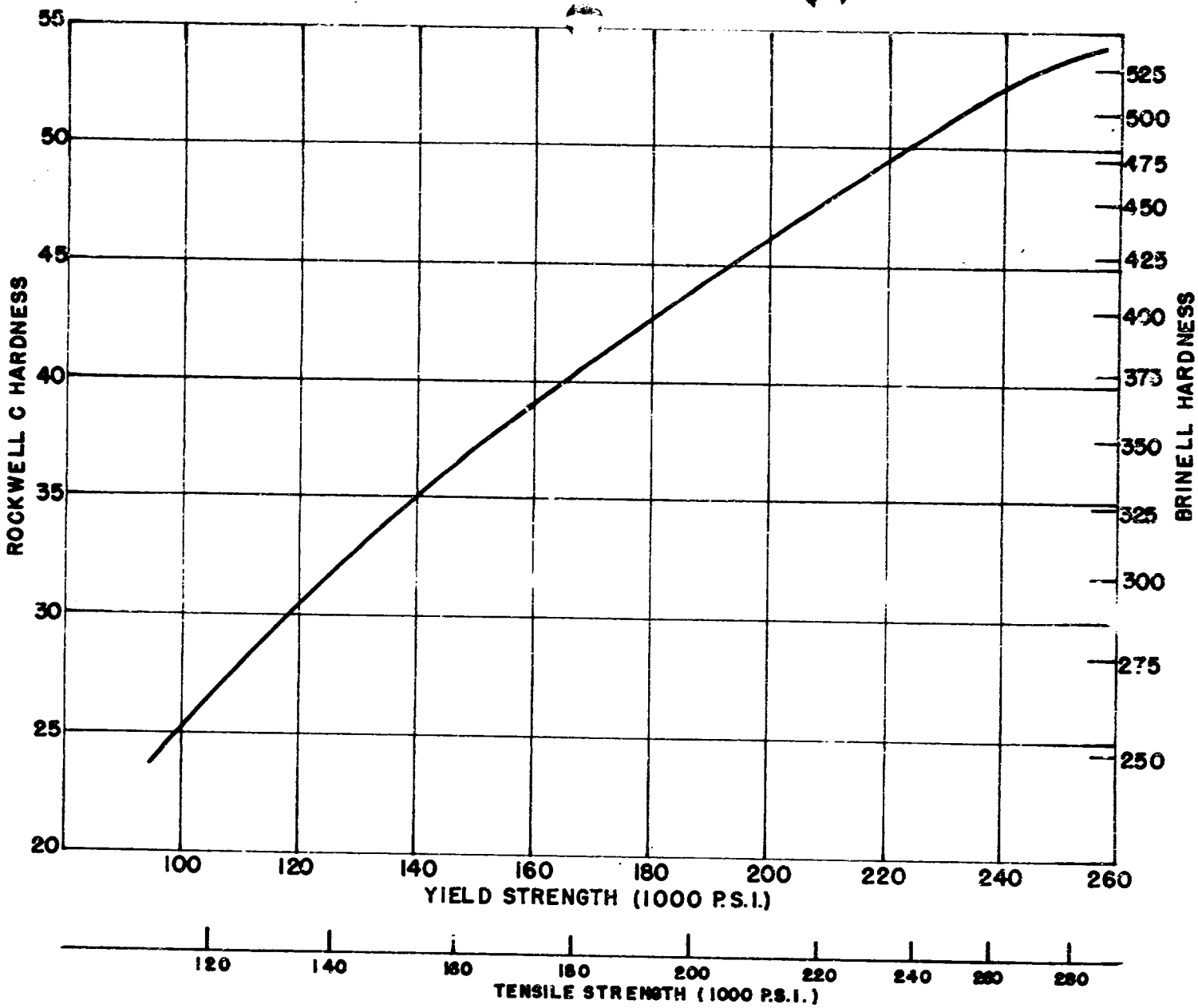


FIGURE 58

RELATIONS OF YIELD & TENSILE STRENGTHS TO ROCKWELL C & BRINELL HARDNESSES

DERIVED FROM FIGURES 56 AND 57.

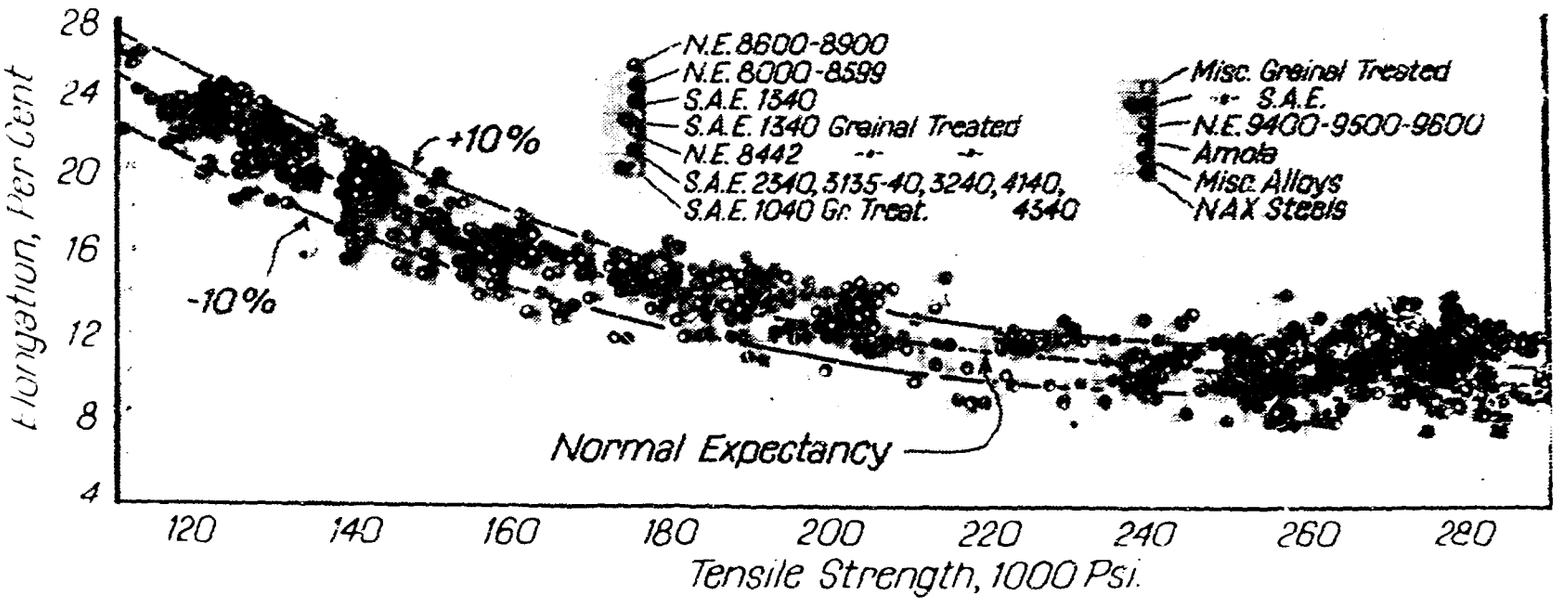
FIGURE 59

RELATION BETWEEN ELONGATION AND TENSILE

STRENGTH FOR TYPICAL COMMERCIAL

STEELS (.25 - .45% CARBON)

(AFTER PATTON(208))


N.E. 8600-8900
N.E. 8000-8599
S.A.E. 1340
S.A.E. 1340 Greinal Treated
N.E. 8442 -"- -"-
S.A.E. 2340, 3135-40, 3240, 4140, 4340
S.A.E. 1040 Gr. Treat.
Misc. Greinal Treated
-"- S.A.E.
N.E. 9400-9500-9600
Amola
Misc. Alloys
NAX Steels
+10%
-10%
Normal Expectancy
Elongation, Per Cent
28
24
20
16
12
8
4
120
140
160
180
200
220
240
260
280
Tensile Strength, 1000 Psi.

WTN.639-7566

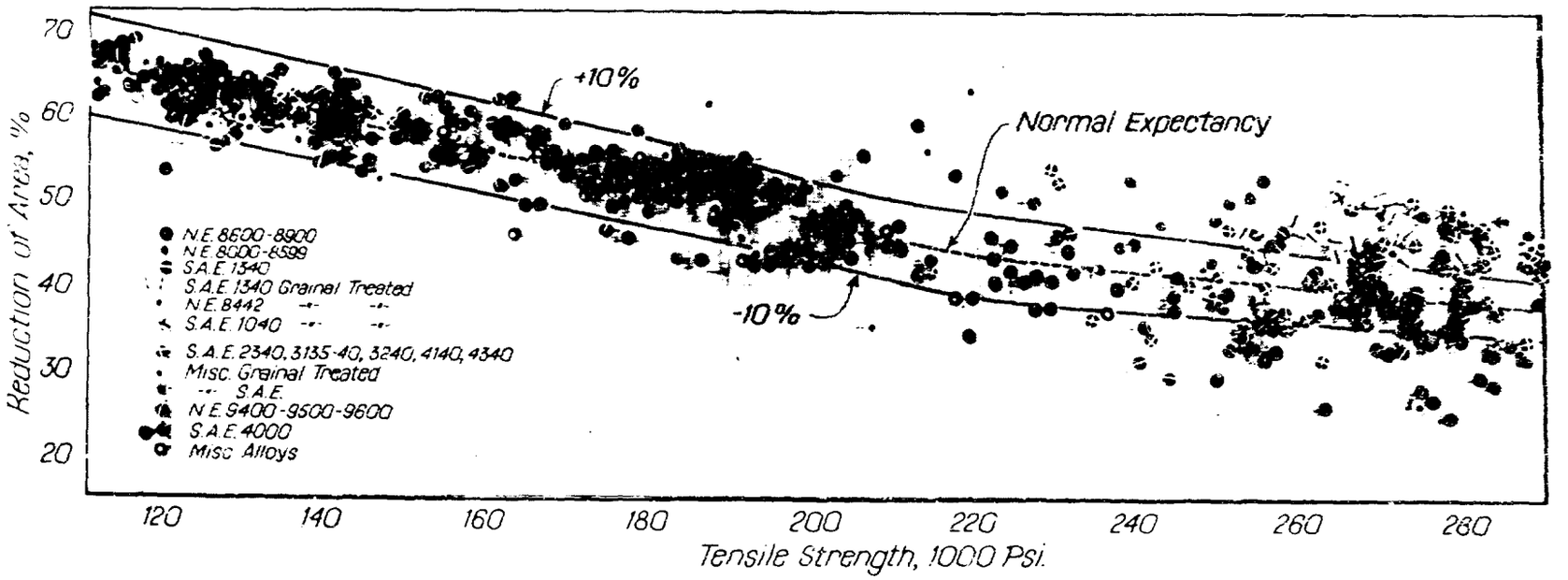
FIGURE 60

RELATION BETWEEN REDUCTION OF AREA

AND TENSILE STRENGTH FOR TYPICAL COMMERCIAL

STEELS (.25 - .45% CARBON)

(AFTER PATTON(208))


Reduction of Area, %
70
60
50
40
30
20
120
140
160
180
200
220
240
260
280
Tensile Strength, 1000 Psi.
+10%
-10%
Normal Expectancy
N.E. 8600-8900
N.E. 8000-8599
S.A.E. 1340
S.A.E. 1340 Grainal Treated
N.E. 8442
S.A.E. 1040
S.A.E. 2340, 3135-40, 3240, 4140, 4340
Misc. Grainal Treated
S.A.E.
N.E. 9400-9500-9600
S.A.E. 4000
Misc Alloys

WTN.639-7569

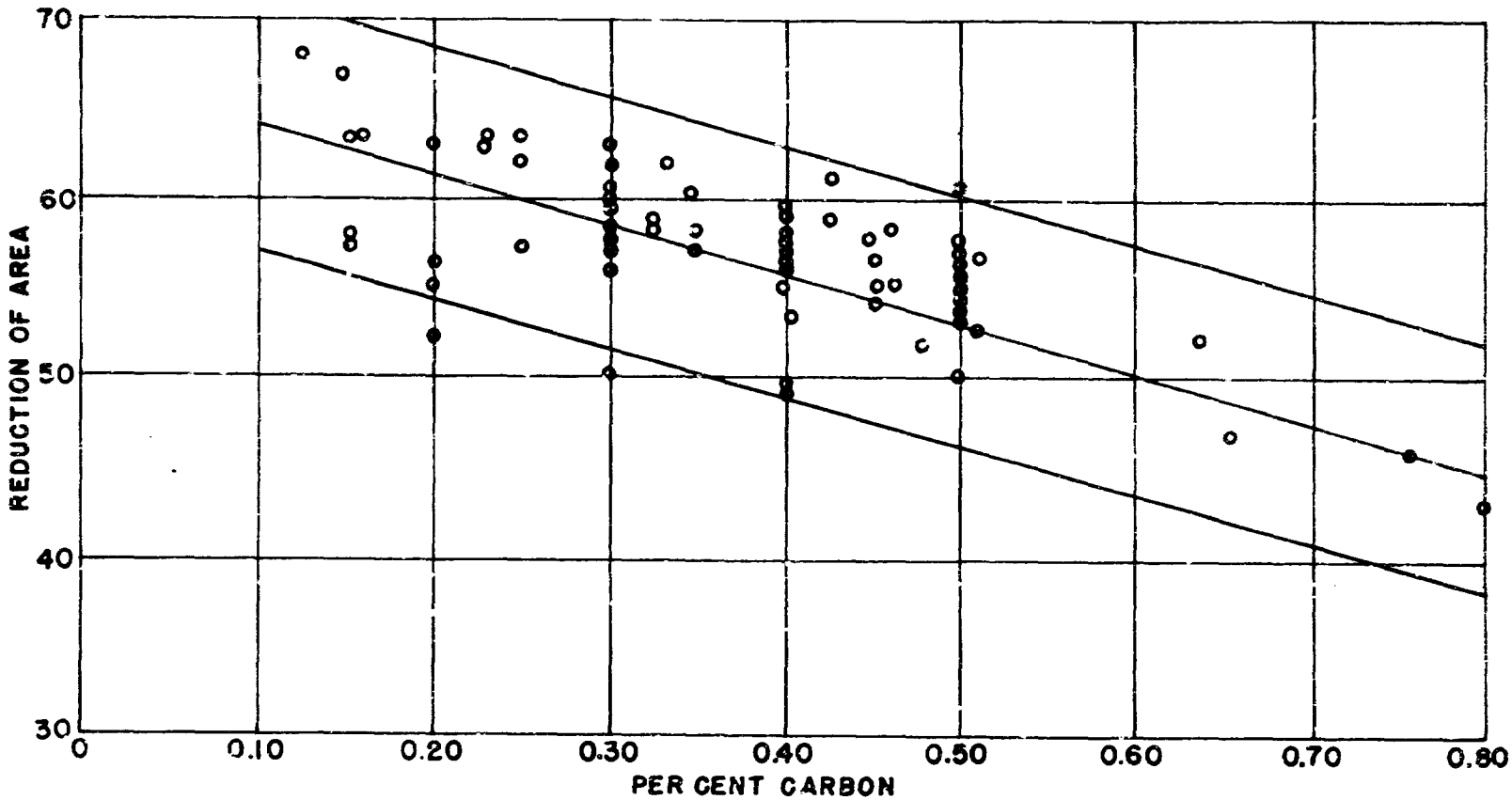


FIGURE 61

EFFECT OF
CARBON CONTENT ON REDUCTION OF AREA OF FULLY HARDENED
& TEMPERED STEELS (TENSILE STRENGTH 150,000 P.S.I.)
(AFTER HAWKES [213])

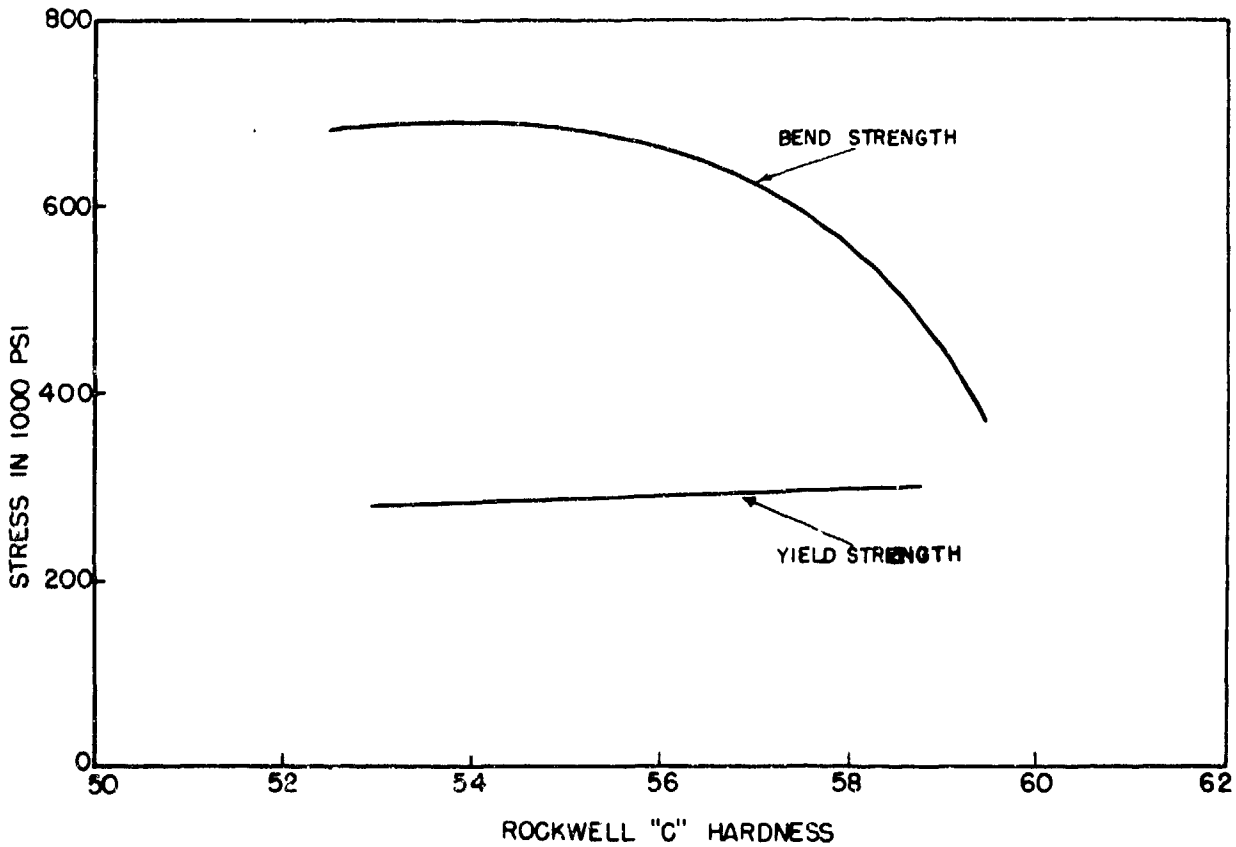


FIGURE 62

EFFECT OF HARDNESS ON YIELD & BEND STRENGTHS OF TEMPERED MARTENSITIC STEELS. (AFTER ZENER & VAN WINKLE[219])

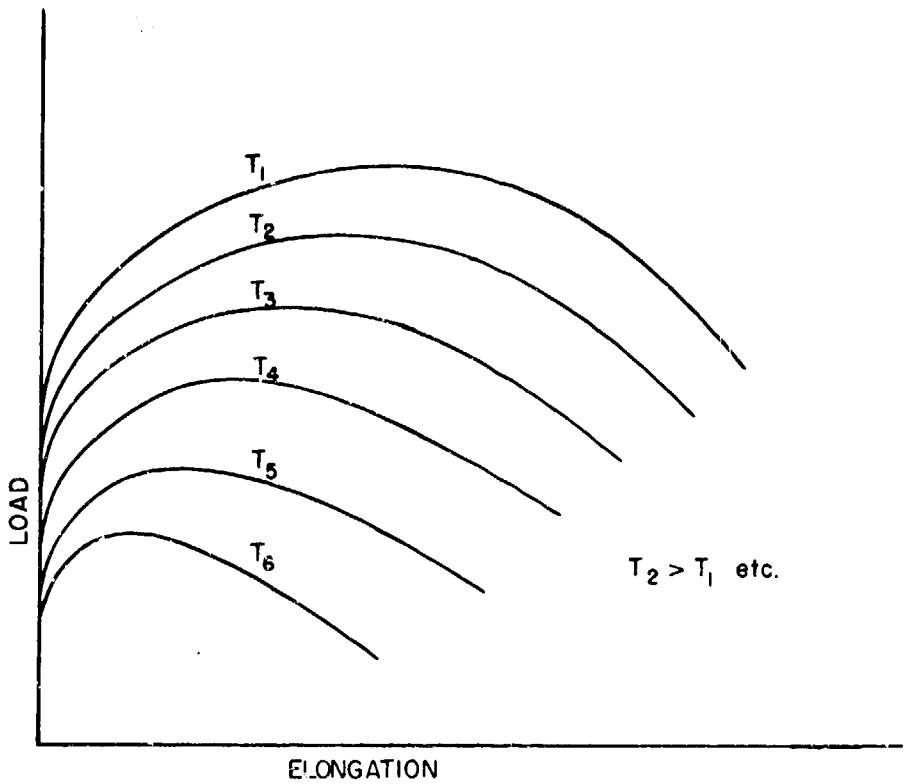


FIGURE 63

EFFECT OF TEMPERATURE ON LOAD-ELONGATION CURVES. (SCHEMATIC)

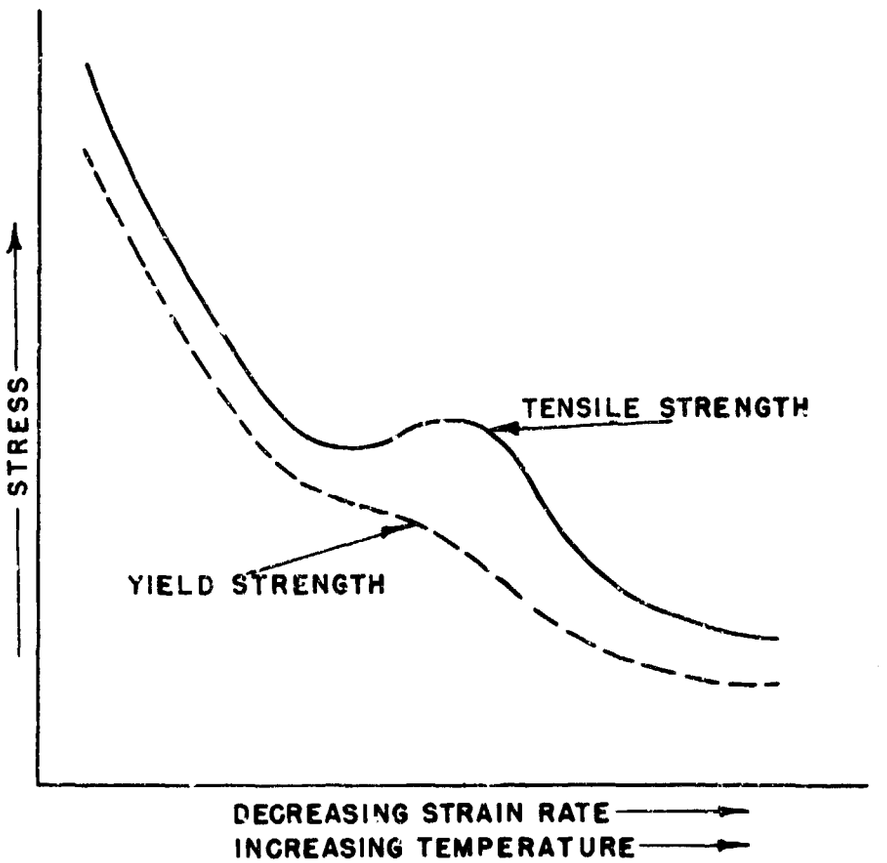


FIGURE 64

EFFECT OF TEMPERATURE & STRAIN RATE ON THE YIELD & TENSILE STRENGTHS OF STEEL (SCHEMATIC)

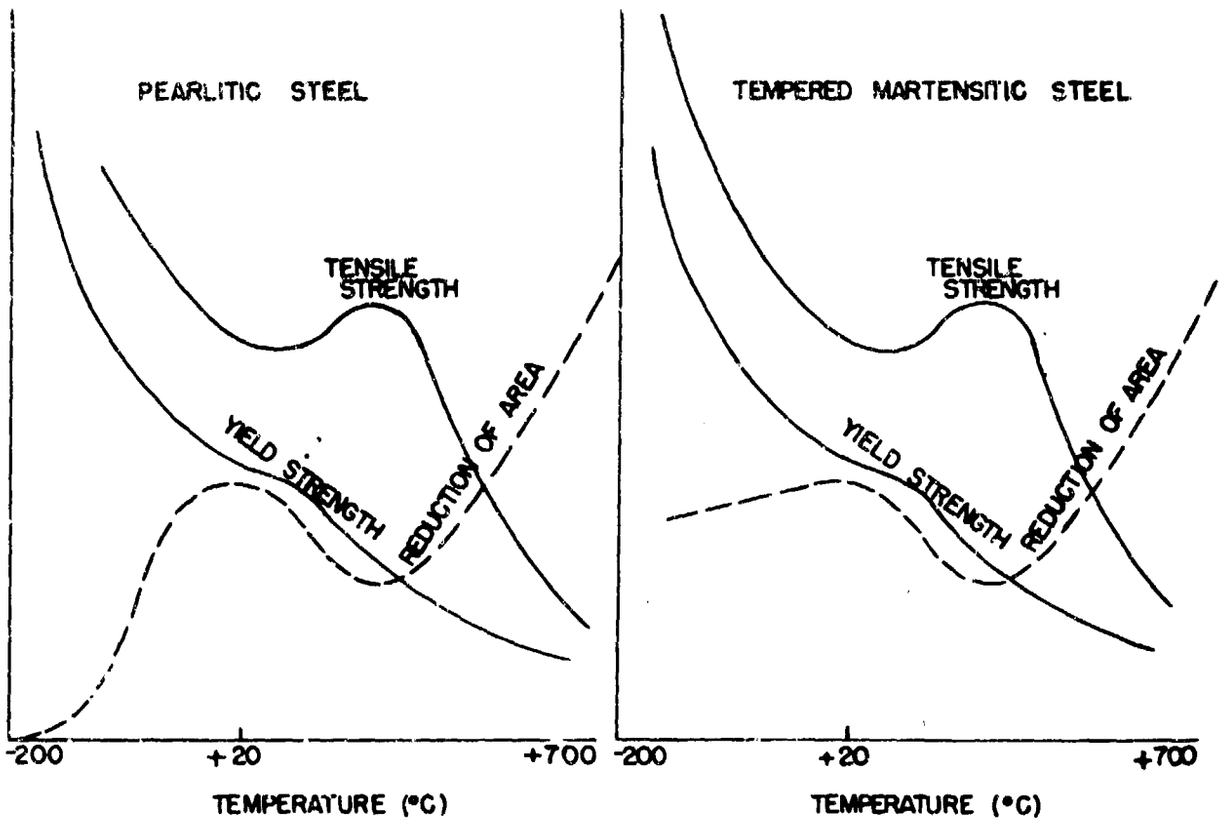
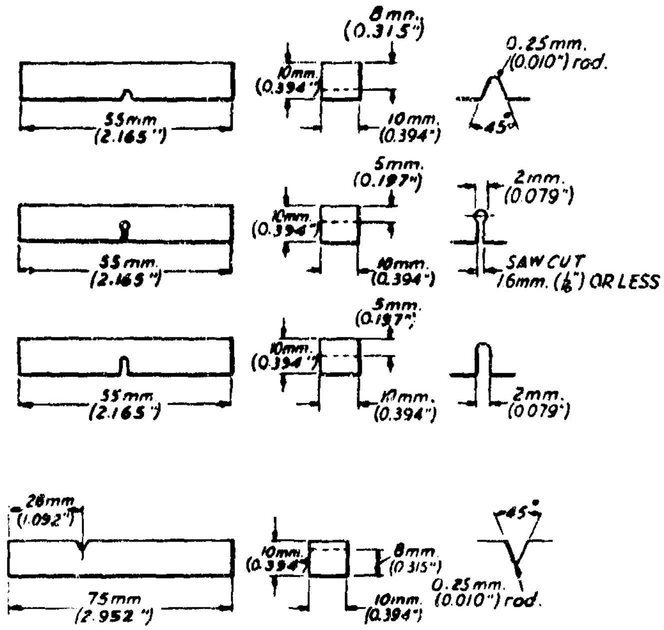


FIGURE 65
EFFECT OF TEMPERATURE ON TENSILE PROPERTIES OF STEELS OF TYPICAL STRUCTURES.

FIGURE 66

TYPICAL NOTCHED-BAR IMPACT SPECIMENS

(FROM A.S.T.M. STANDARDS(221))



PERMISSIBLE VARIATIONS:

CROSS-SECTION DIMENSIONS	±0.025mm.(0.001")
LENGTH OF SPECIMEN	±0.25mm.(0.010")
ANGLE OF NOTCH	±1°

WTN.630-7564

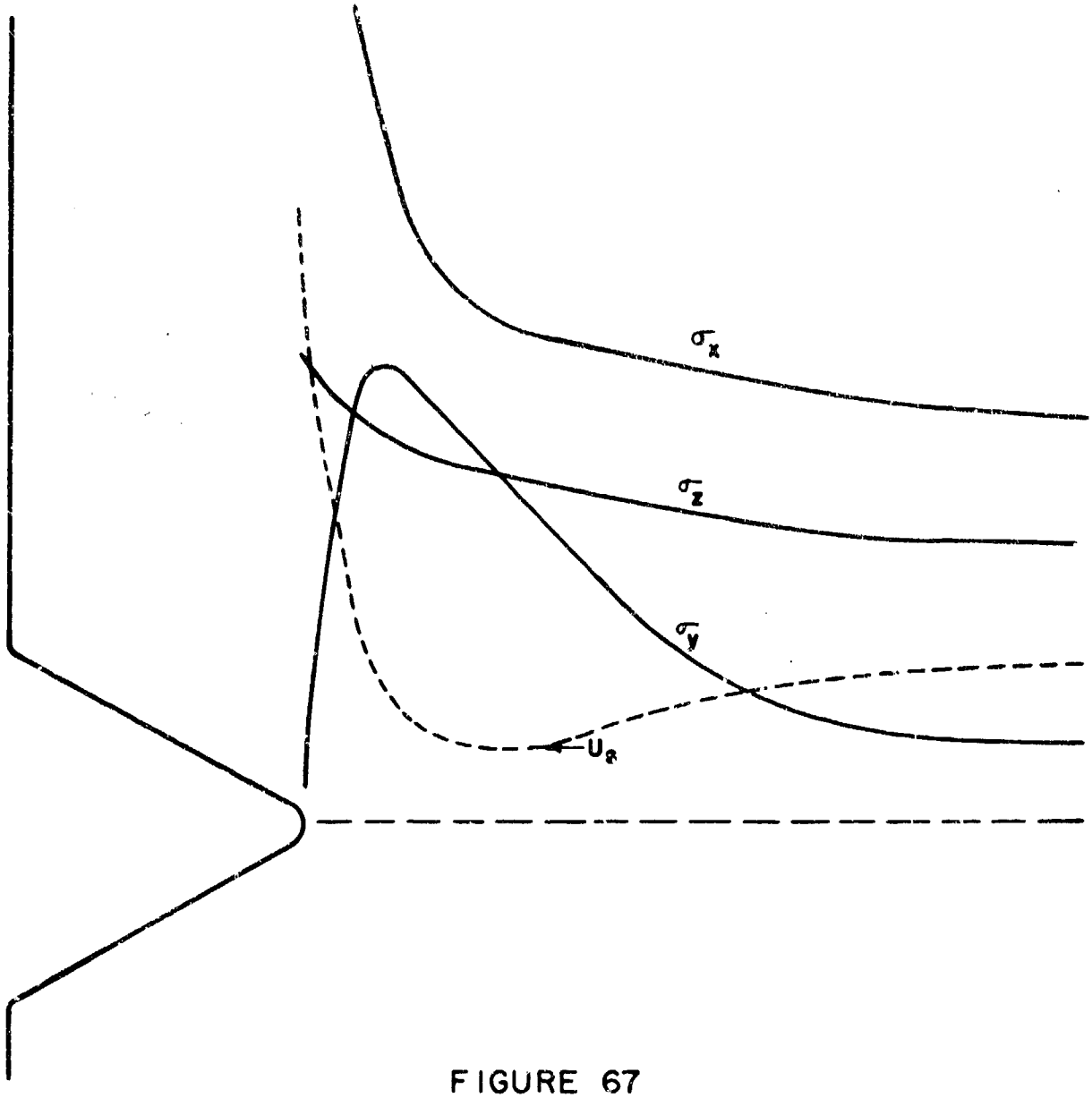


FIGURE 67

VARIATION OF STATE OF STRESS ACROSS A NOTCHED CYLINDER IN TENSION. (AFTER GENSAMER[224]).

σ_x = LONGITUDINAL STRESS
σ_y = RADIAL STRESS
σ_z = TRANSVERSE (OR CIRCUMFERENTIAL) STRESS
U_s = SHEAR STRAIN ENERGY

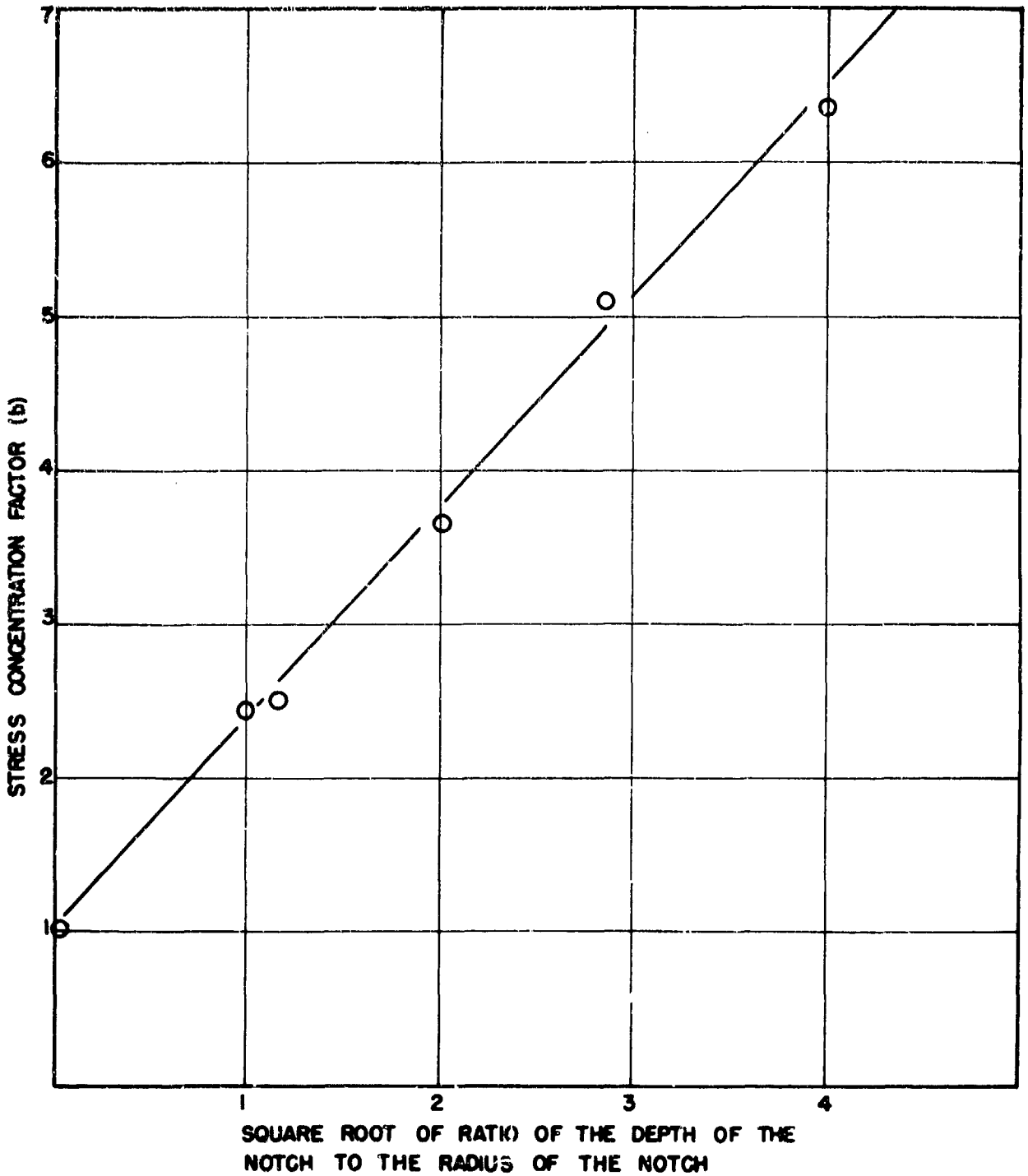


FIGURE 68

EFFECT OF NOTCH DEPTH & SHARPNESS ON THE LONGITUDINAL TENSILE STRESS AT THE BASE OF A NOTCH (AFTER COKER & FILON[225]) (FOR "V" NOTCH CHARPY BAR)

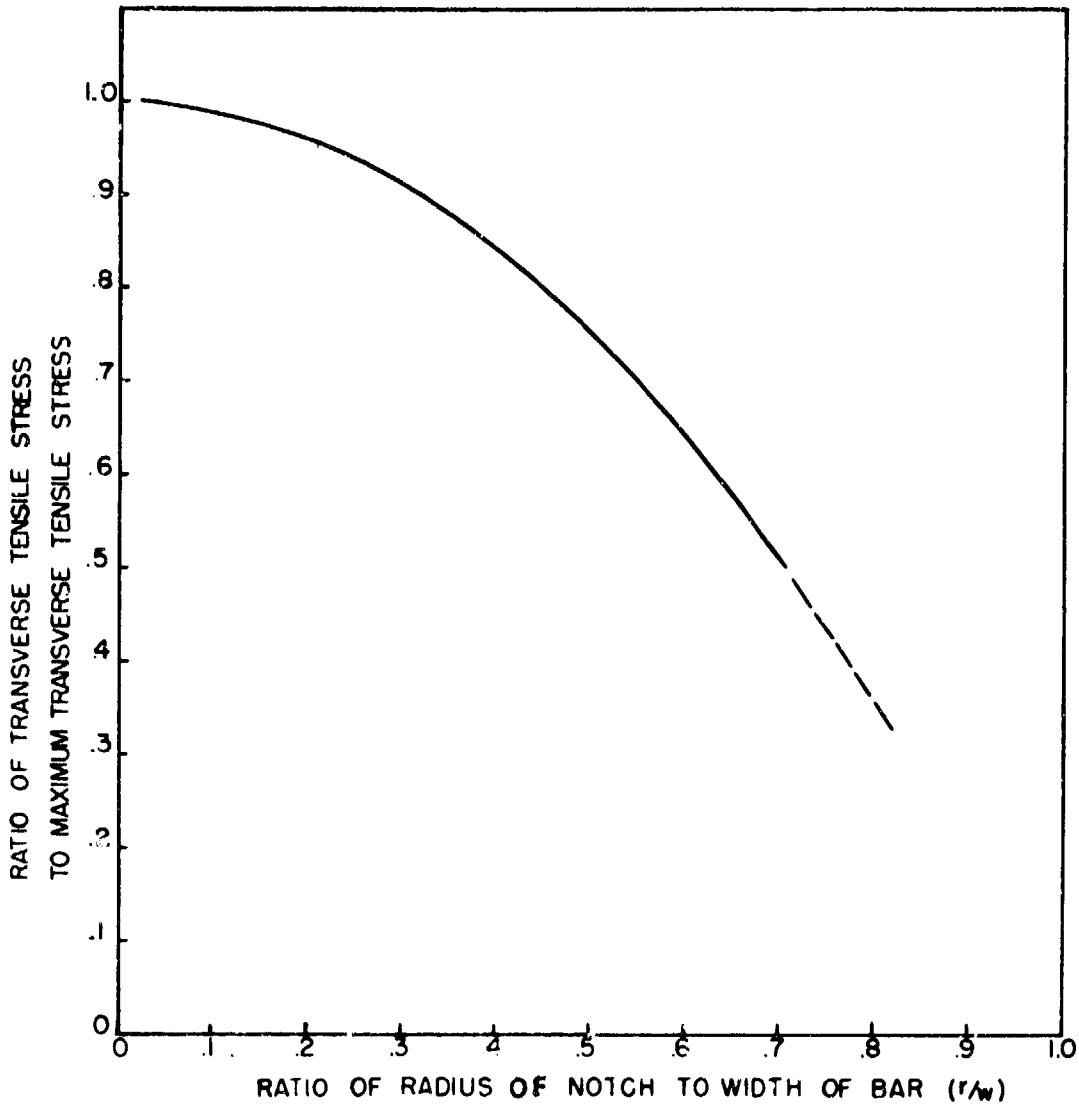


FIGURE 69

EFFECT OF DIMENSIONS OF NOTCHED BAR ON THE TRANSVERSE CONSTRAINT (AFTER HOLLOMON[191])

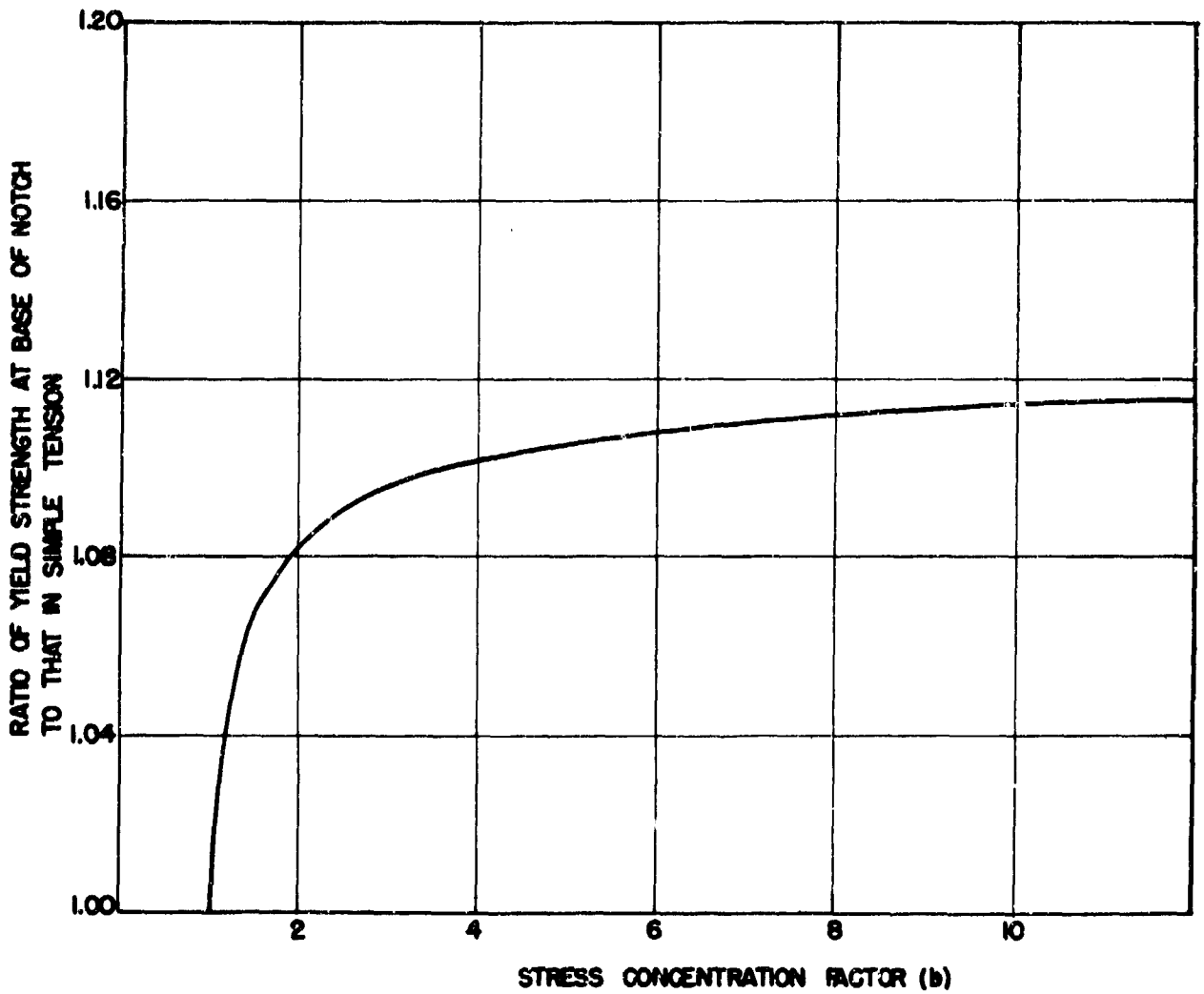


FIGURE 70

EFFECT OF STRESS CONCENTRATION FACTOR ON TENSILE STRESS NECESSARY FOR YIELDING (V-NOTCH CHARPY BAR) (AFTER HOLLOMON[191])

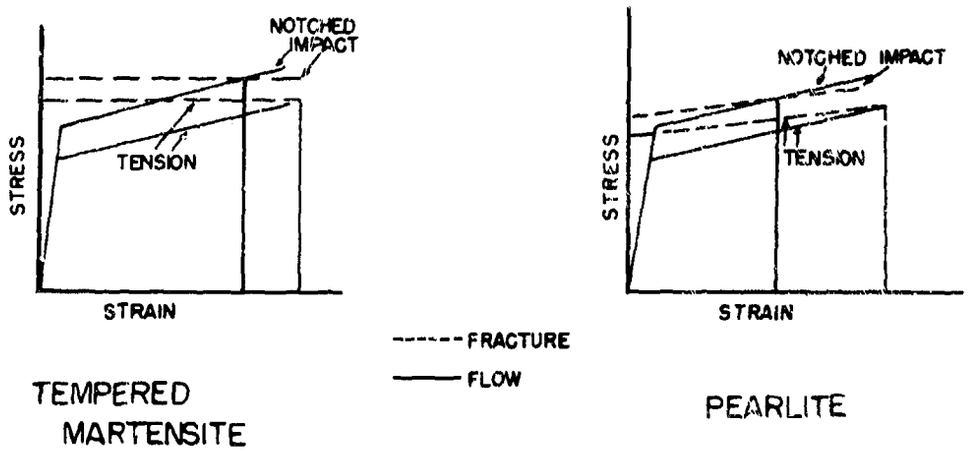
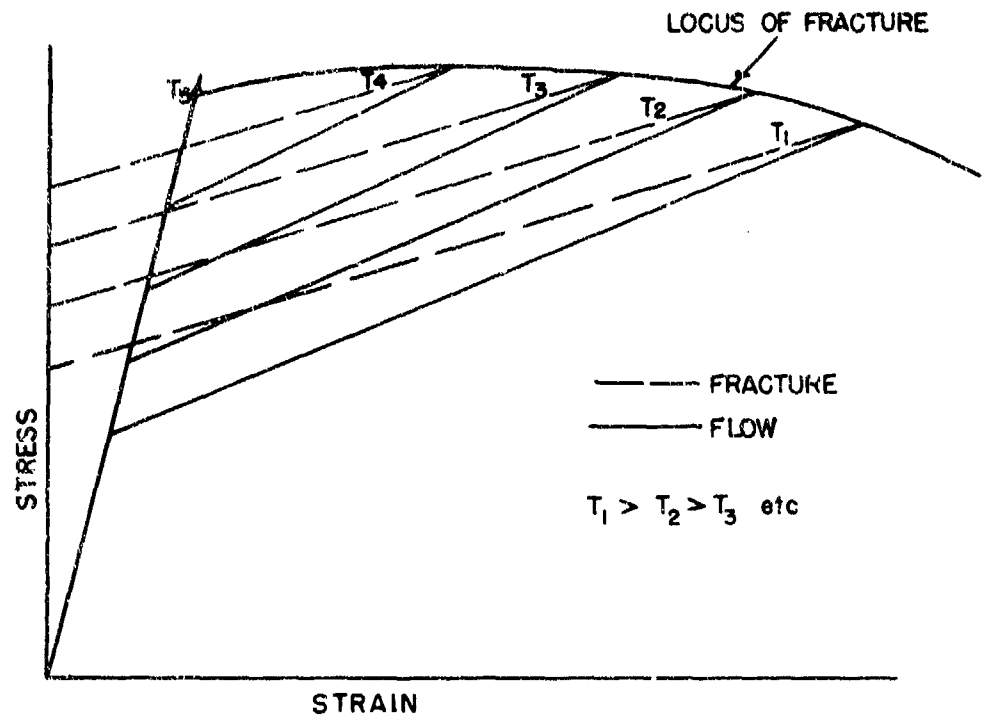


FIGURE 71

EFFECT OF NOTCH AND IMPACT ON STRESS-STRAIN CURVES OF TYPICAL STEELS. (SCHEMATIC)



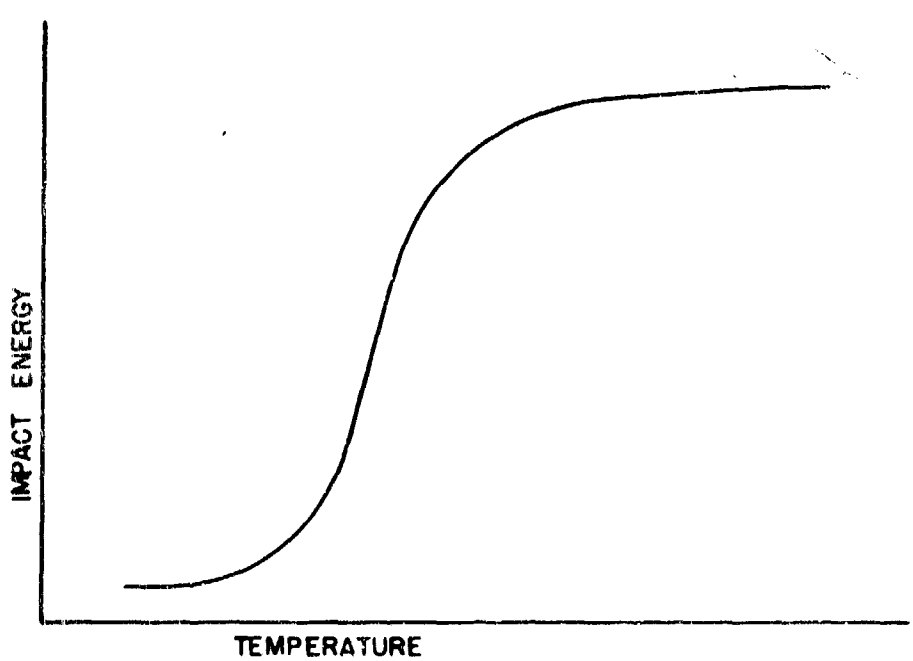


FIGURE 72

STRESS - STRAIN CURVES & IMPACT ENERGY AT LOW TEMPERATURES OF A PEARLITIC STEEL.

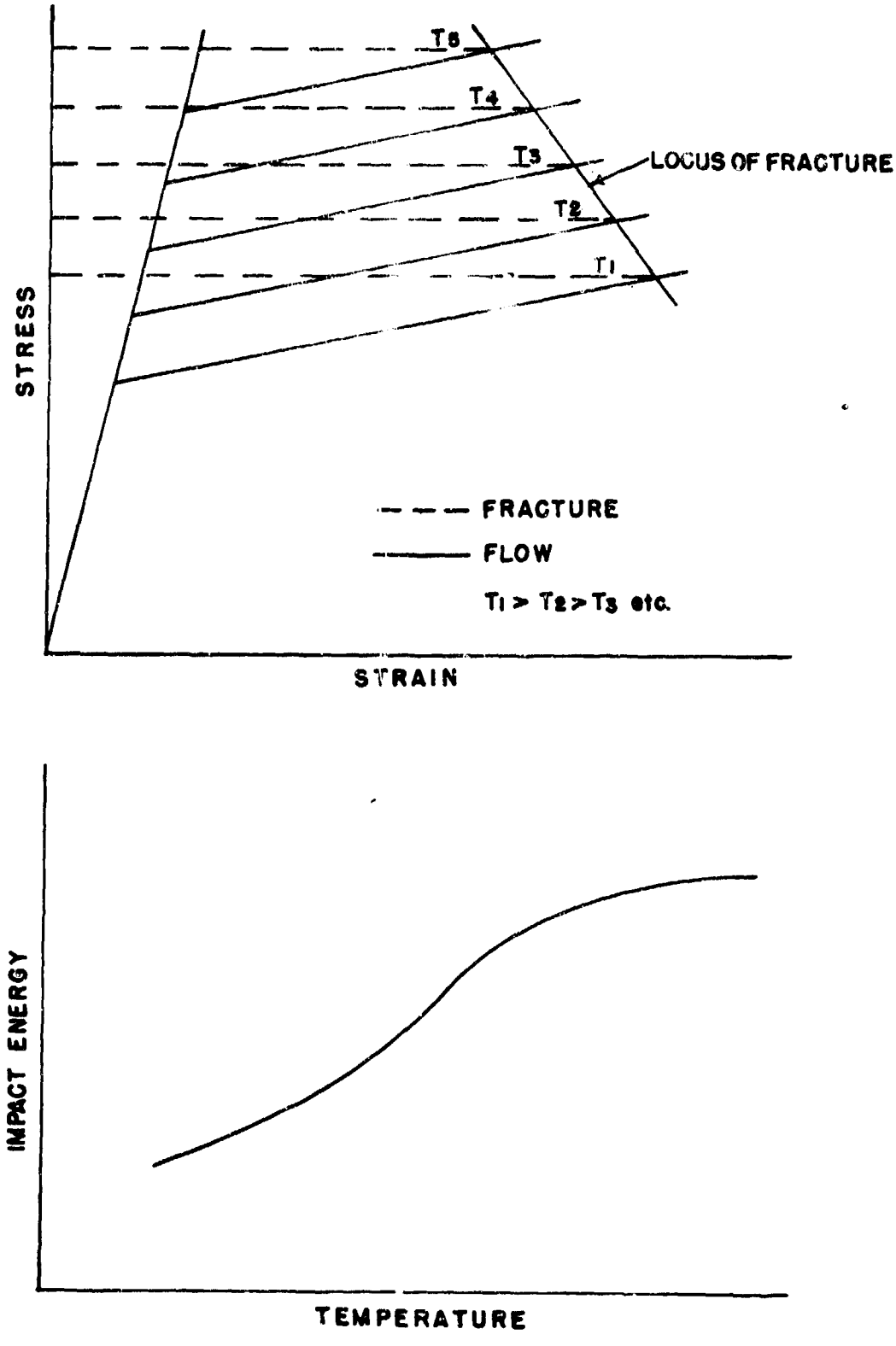


FIGURE 73

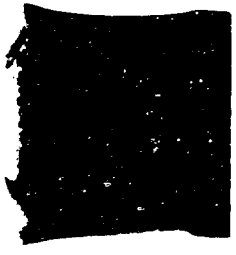
STRESS-STRAIN CURVES & IMPACT ENERGY AT LOW TEMPERATURES OF A TEMPERED MARTENSITIC STEEL

FIGURE 74

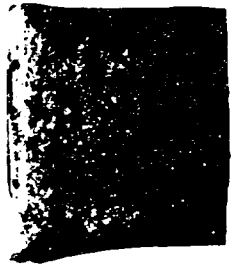
TYPICAL NOTCHED-BAR FRACTURES

OF STEEL

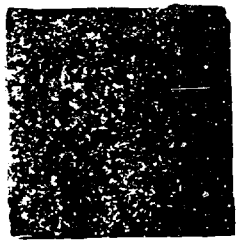
FIBROUS



BRIGHT CRYSTALLINE AND FIBROUS



100% BRIGHT CRYSTALLINE



FIBROUS

DULL CRYSTALLINE AND FIBROUS

100% DULL CRYSTALLINE

WTN.710-2383

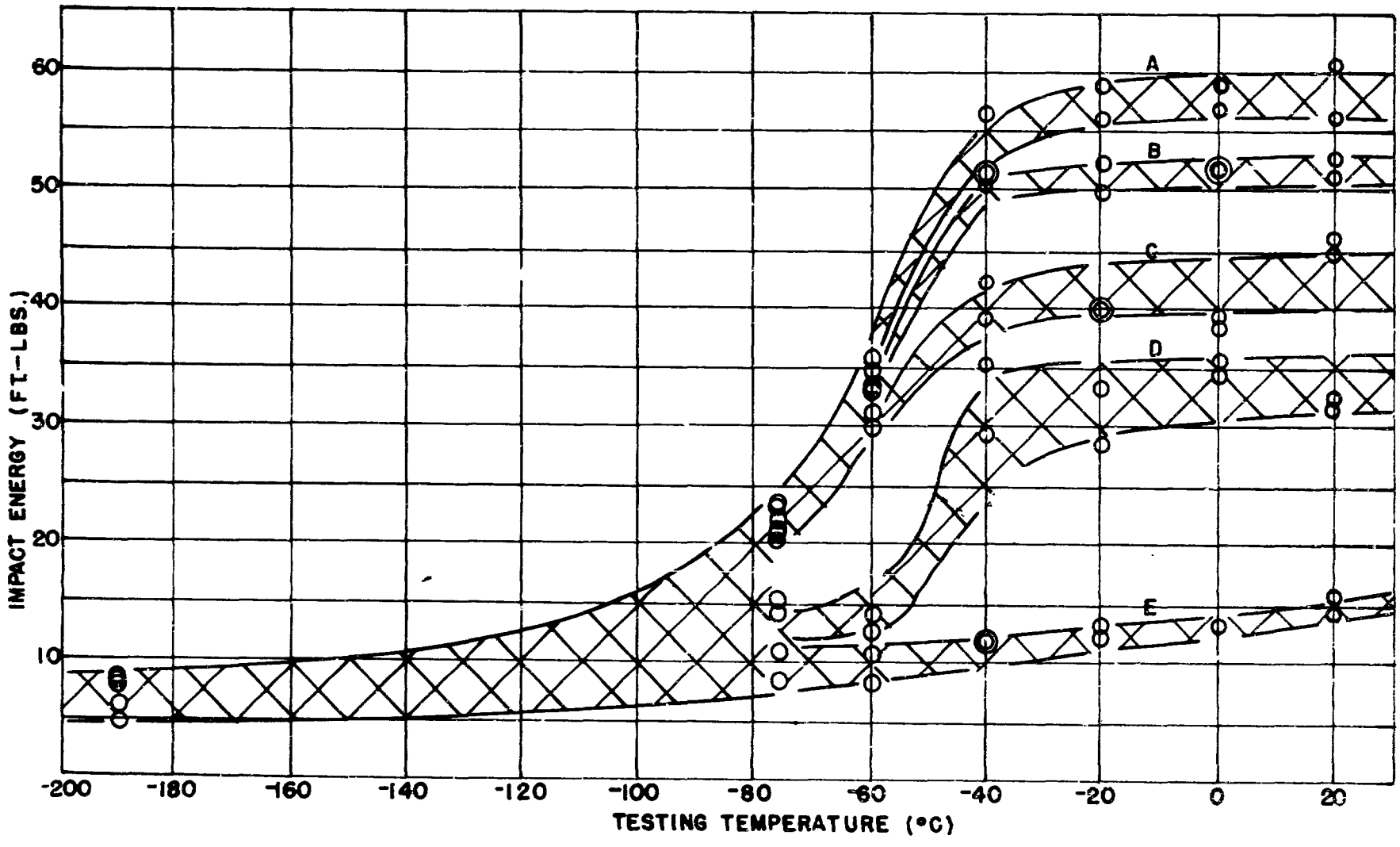


FIGURE 75

NOTCHED BAR IMPACT ENERGY AT LOW TEMPERATURE FOR A TEMPERED MARTENSITIC STEEL AT VARIOUS YIELD STRENGTHS (YIELD STRENGTH A<B etc.)

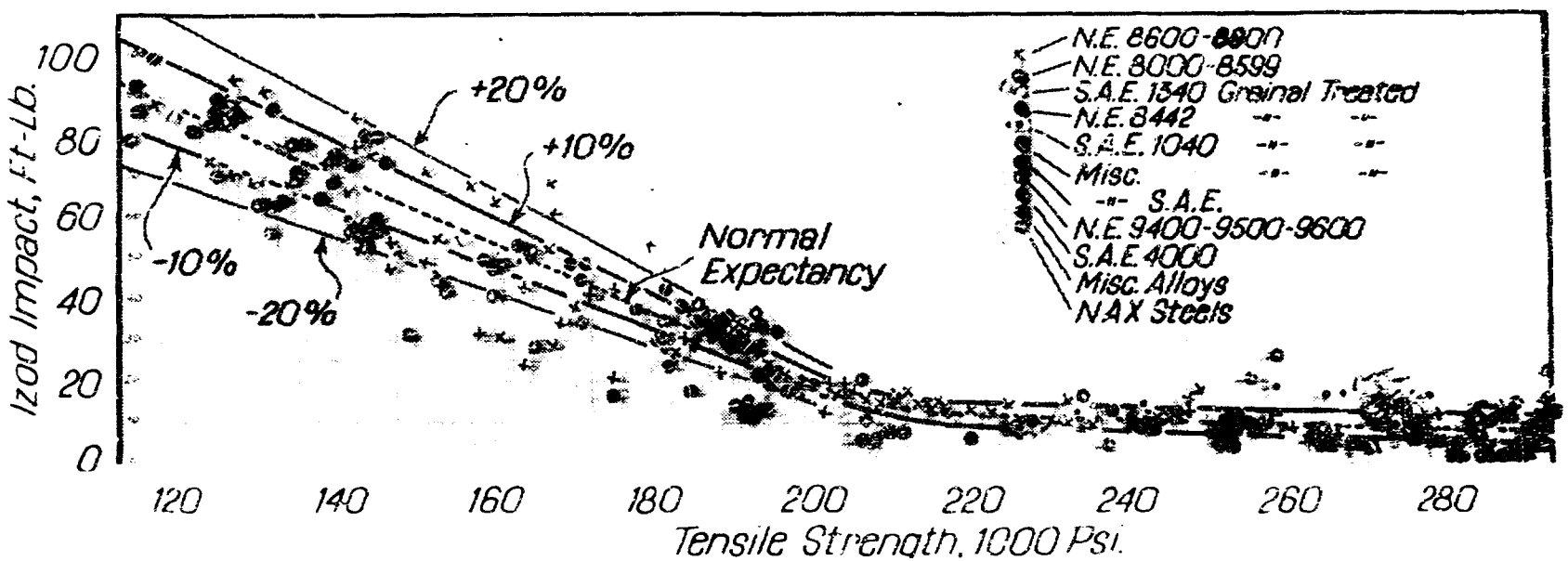
FIGURE 76

VARIATION OF IZOD IMPACT ENERGY WITH

TENSILE STRENGTH OF COMMERCIAL STEELS

(ROOM TEMPERATURE TESTS, .25 - .45% CARBON)

(AFTER PATTON(208))


Izod Impact, Ft-Lb.
100
80
60
40
20
0
120
140
160
180
200
220
240
260
280
Tensile Strength, 1000 Psi.
+20%
+10%
-10%
-20%
Normal Expectancy
N.E. 8600-8800
N.E. 8000-8599
S.A.E. 1340 Greinal Treated
N.E. 8442
S.A.E. 1040
Misc.
S.A.E.
N.E. 9400-9500-9600
S.A.E. 4000
Misc. Alloys
NAX Steels

WTN.639-7567

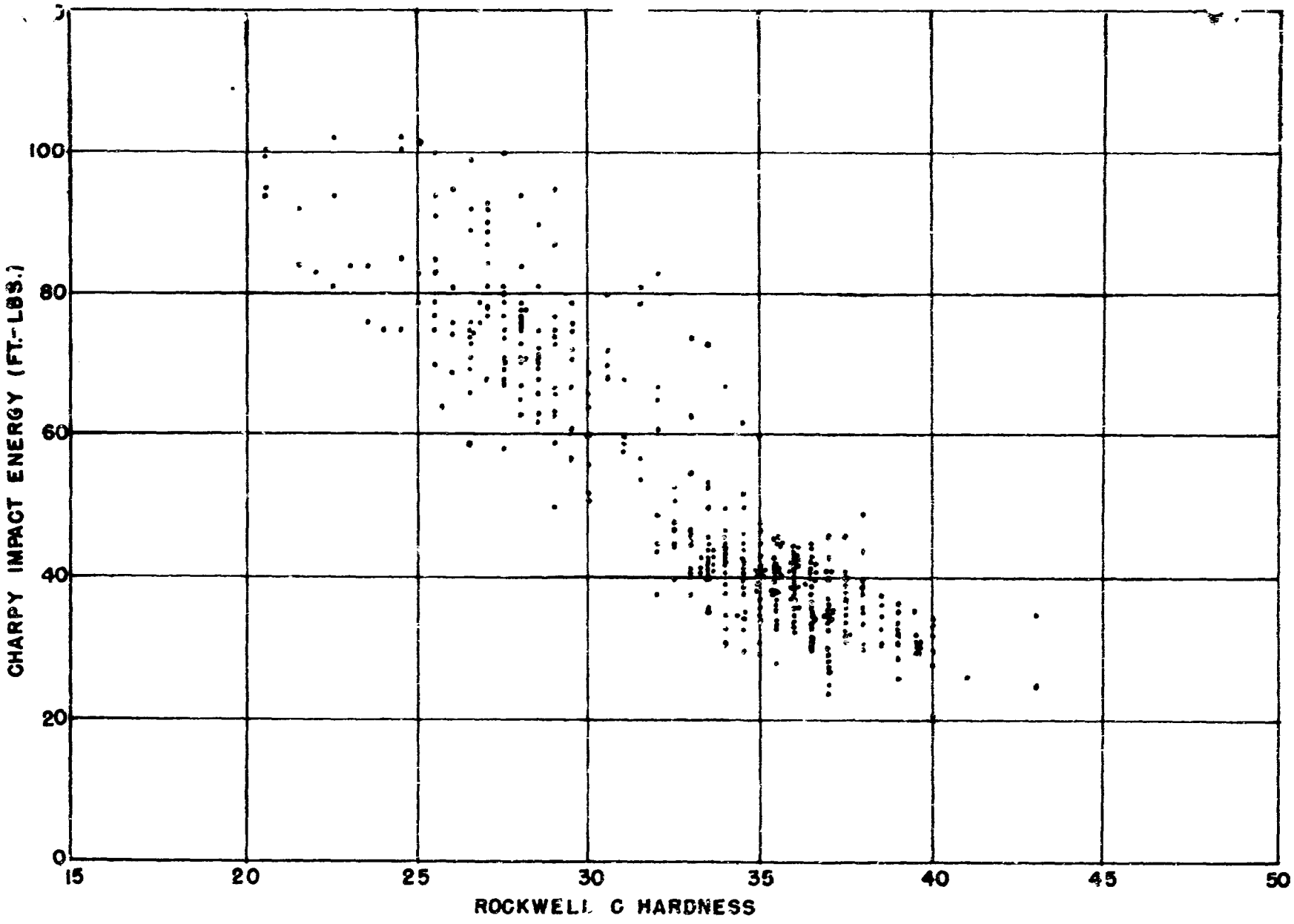


FIGURE 77

VARIATION OF IMPACT ENERGY WITH HARDNESS FOR FULLY HARDENED STEELS (SPECIMENS TAKEN PARALLEL TO PRINCIPAL ROLLING DIRECTION OF PLATES). ROOM TEMPERATURE TESTS .30-.40% CARBON.

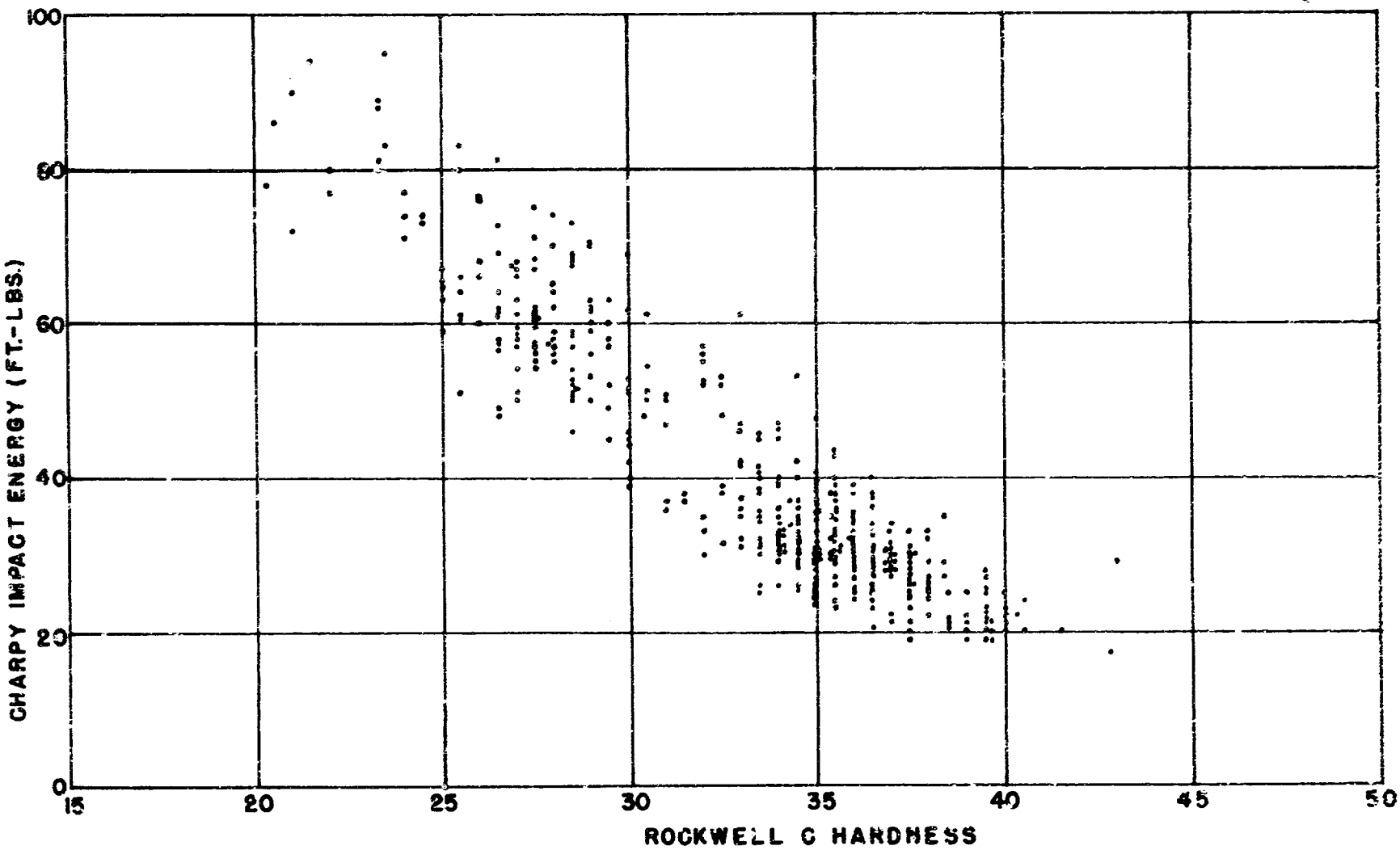


FIGURE 78

VARIATION OF IMPACT ENERGY WITH HARDNESS FOR FULLY HARDENED STEELS (SPECIMENS TAKEN TRANSVERSE TO PRINCIPAL ROLLING DIRECTION OF PLATES).

ROOM TEMPERATURE TESTS .30-.40 % CARBON.

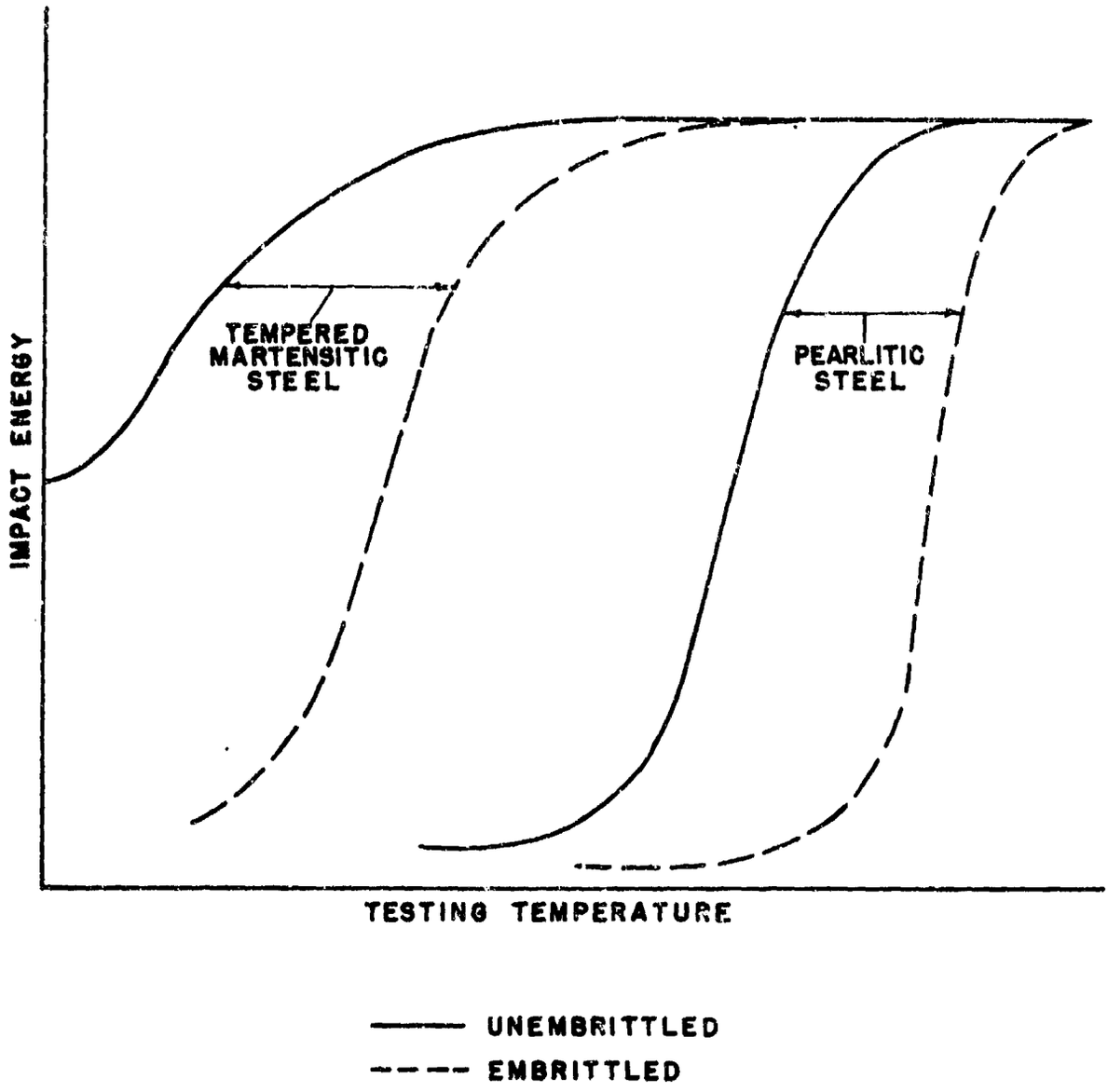


FIGURE 79

EFFECT OF TEMPER BRITTLENESS ON THE IMPACT CURVES OF STEELS OF TYPICAL STRUCTURE. (SCHEMATIC)

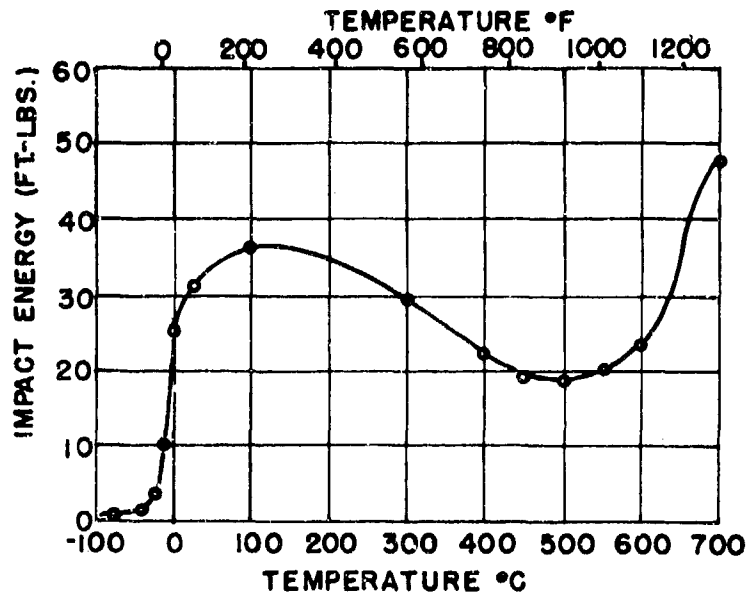


FIGURE 80

IMPACT ENERGY AS A FUNCTION OF TEMPERATURE FOR TYPICAL LOW CARBON PEARLITIC STEEL. (AFTER EPSTEIN[36])

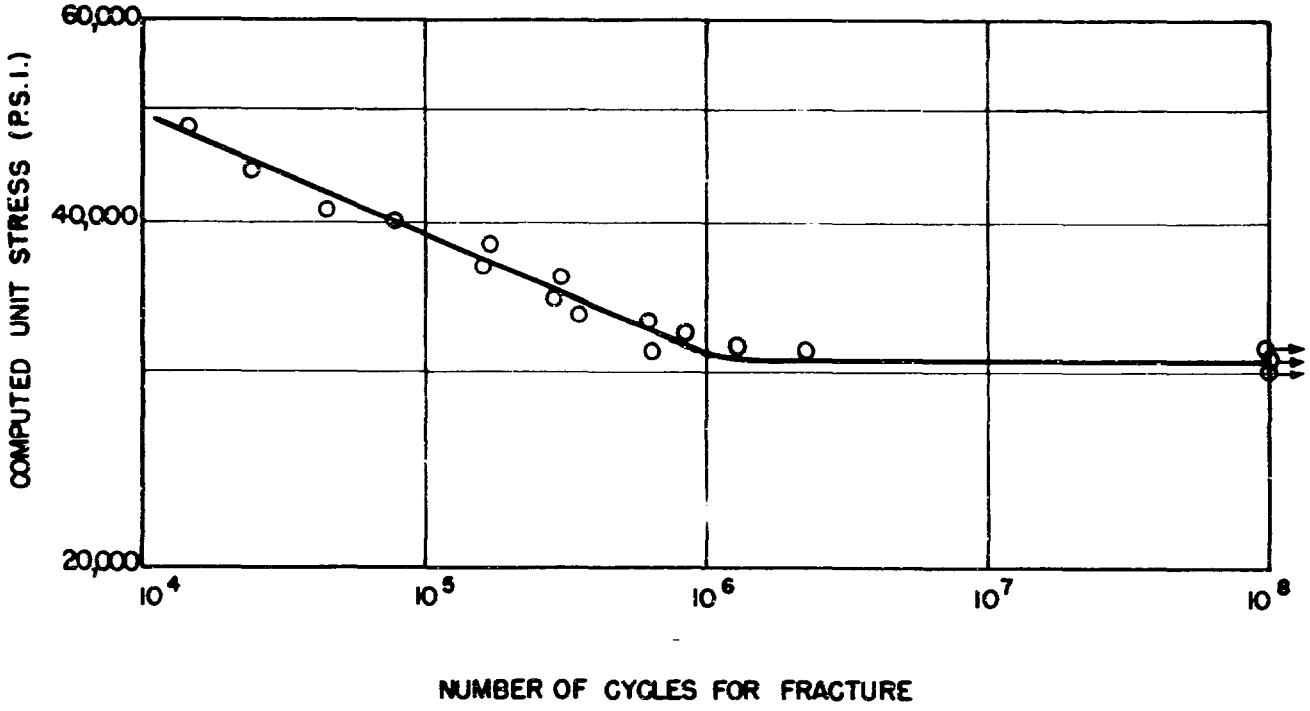


FIGURE 81

TYPICAL LOGARITHMIC S-N CURVE (AFTER MOORE & KOMMERS[232])

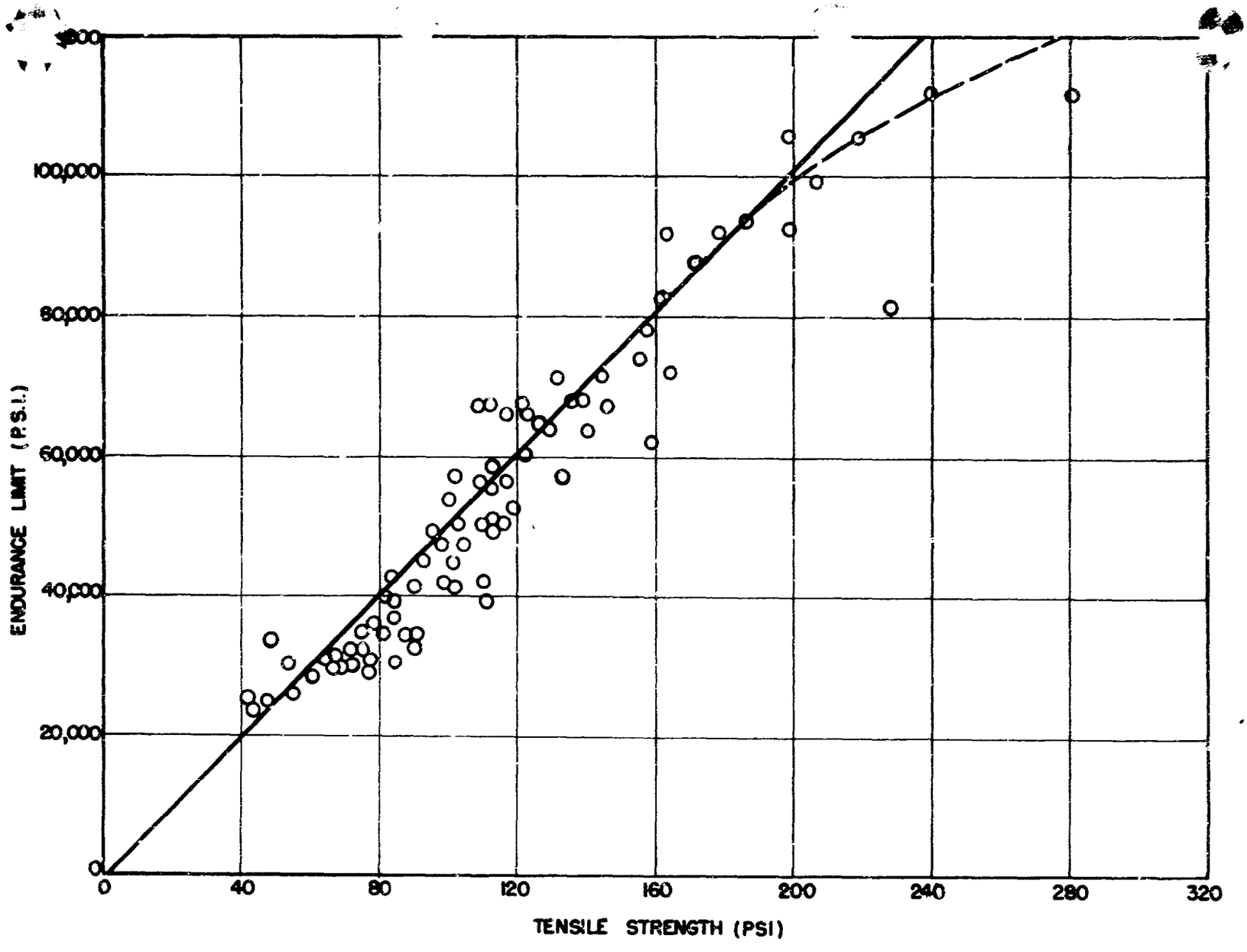


FIGURE 82
RELATION BETWEEN ENDURANCE LIMIT & TENSILE STRENGTH OF STEELS
(AFTER MOORE & KOMMERS[232])

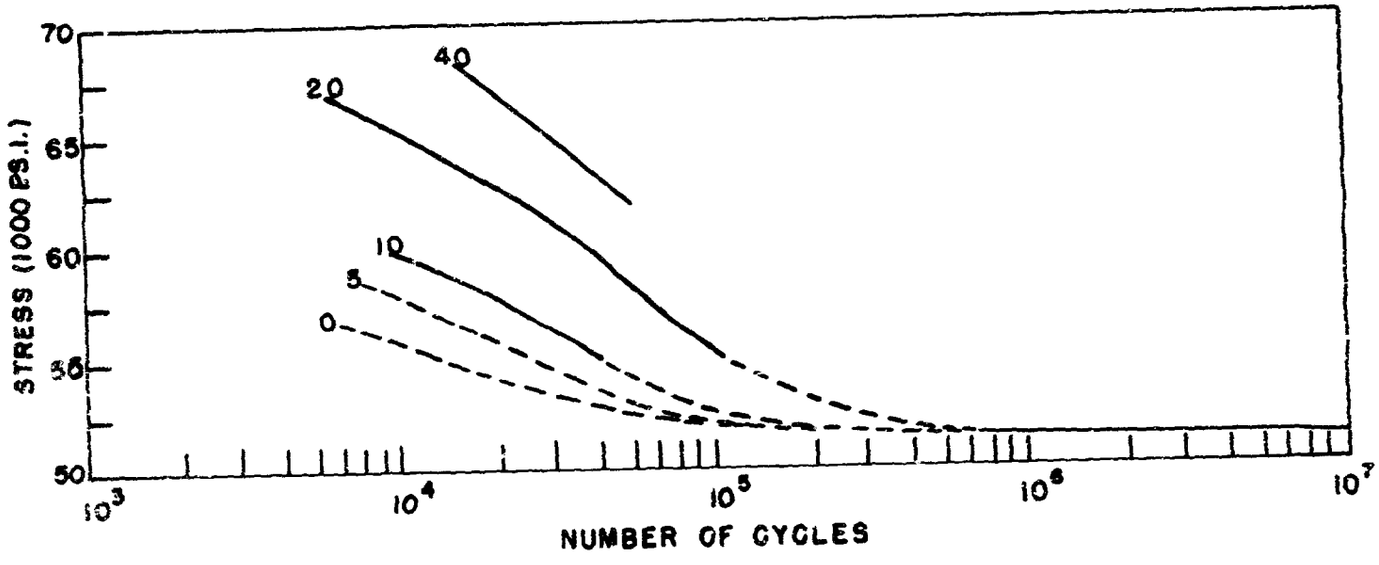
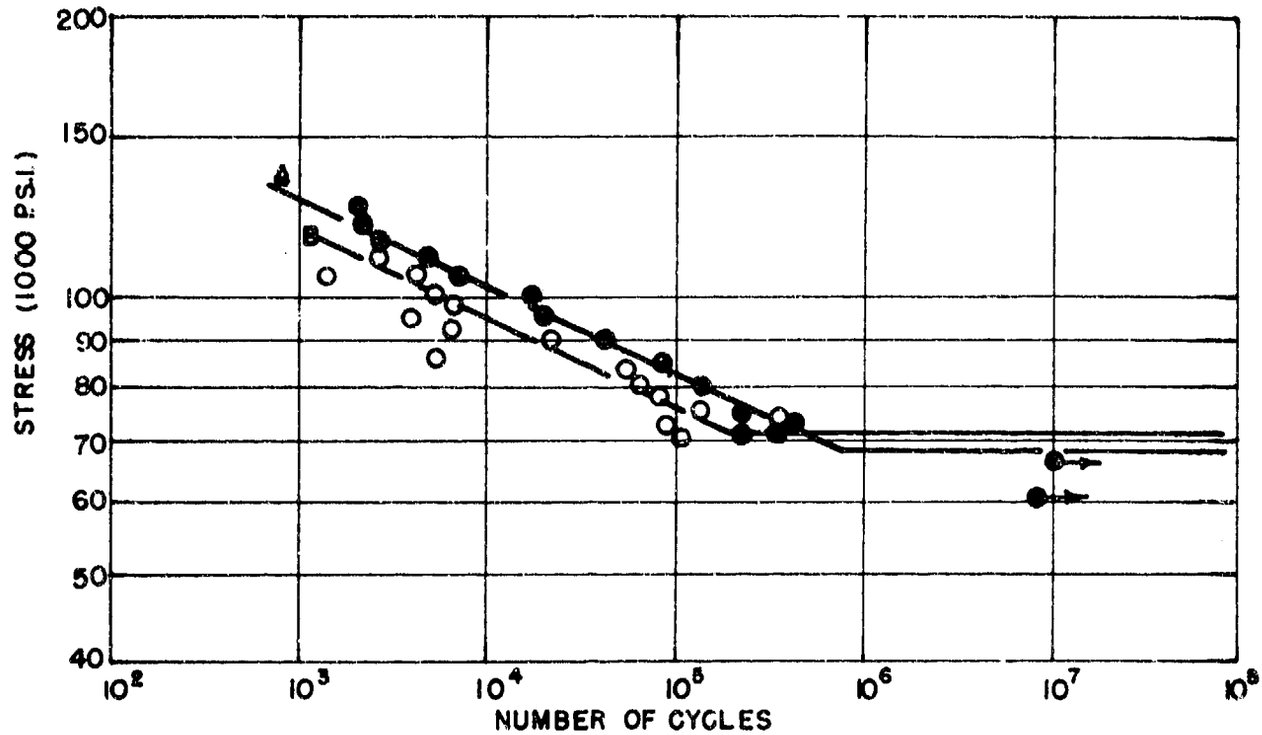


FIGURE 83

EFFECT OF OVER STRESSING IN REDUCING THE ENDURANCE LIMIT
(NUMERALS REPRESENT PERCENTAGE OF DECREASE)
(FROM DATA OF KOMMERS[237]
PLOTTED BY SANFORD[238])



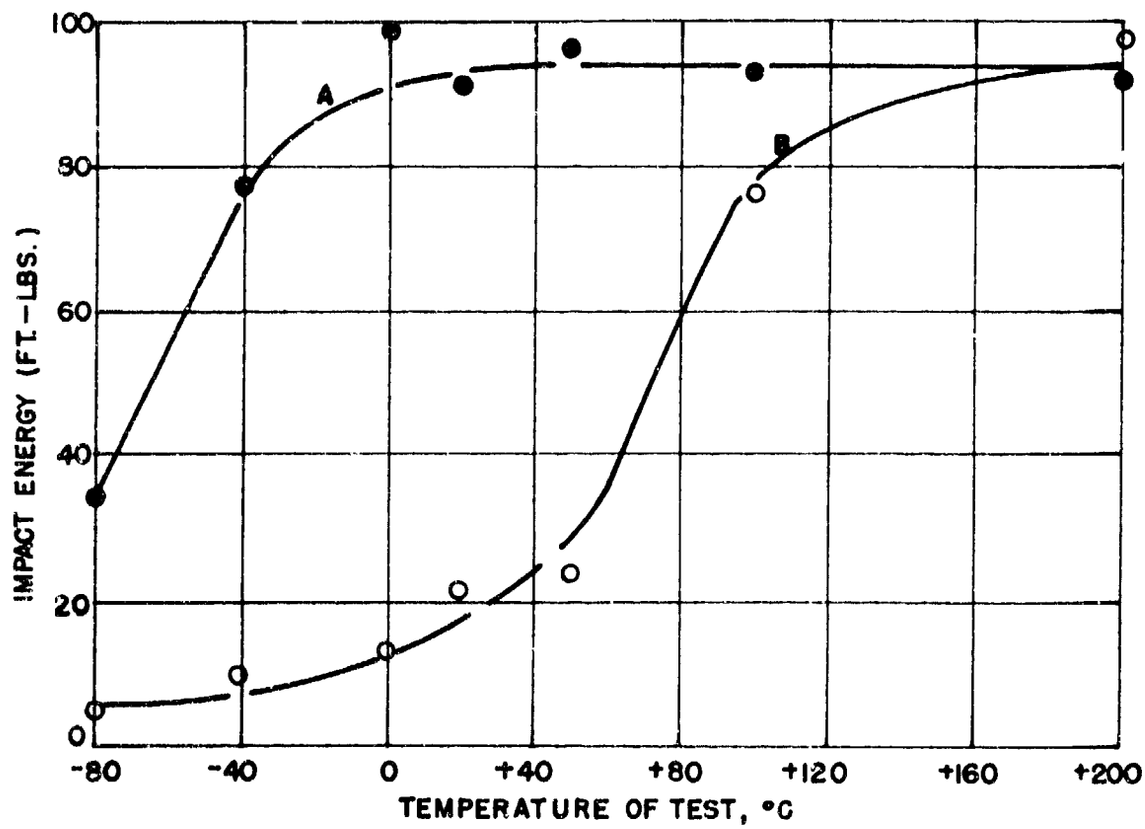
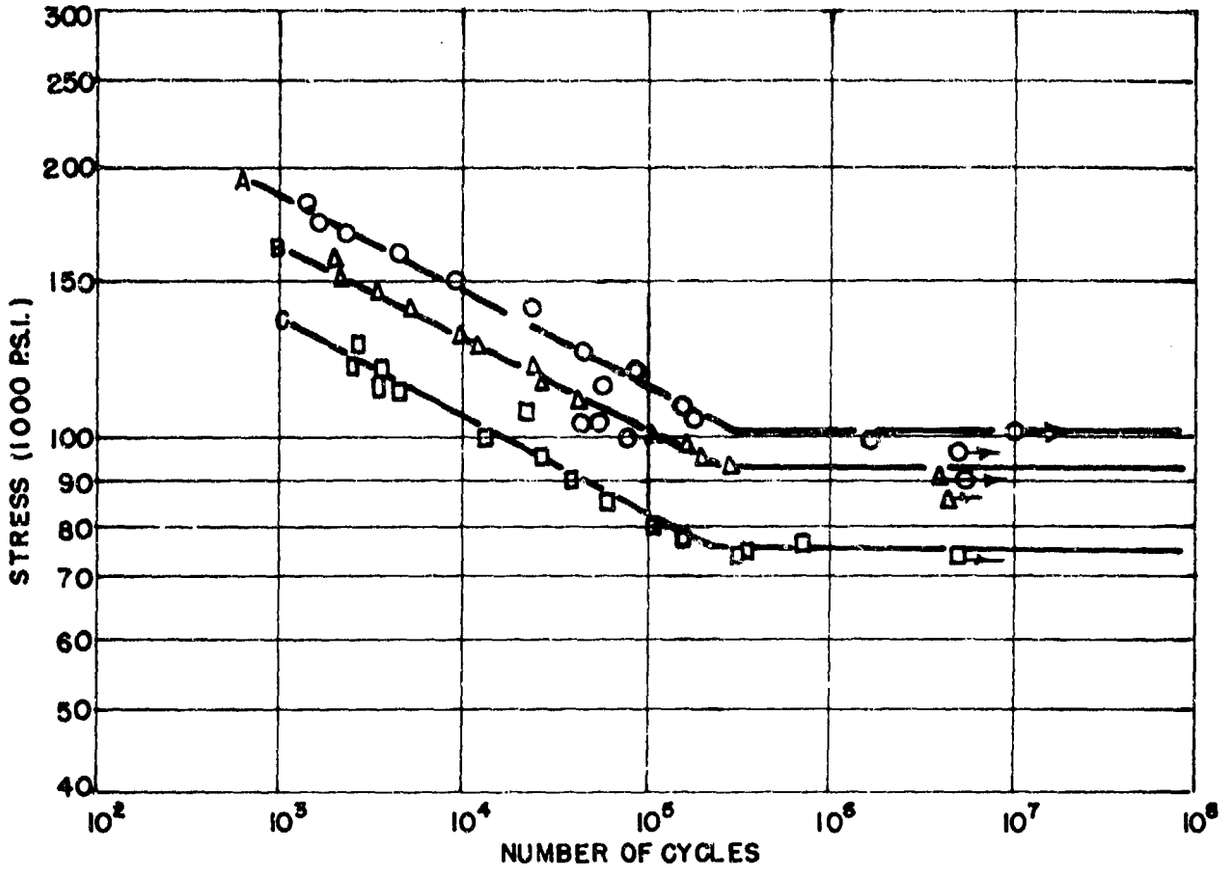


FIGURE 84

EFFECT OF VARYING DEGREES OF TEMPER BRITTLENESS ON THE FATIGUE & IMPACT PROPERTIES OF SAE 3130 STEEL.

(AFTER HOLLOMON ETAL.[240])



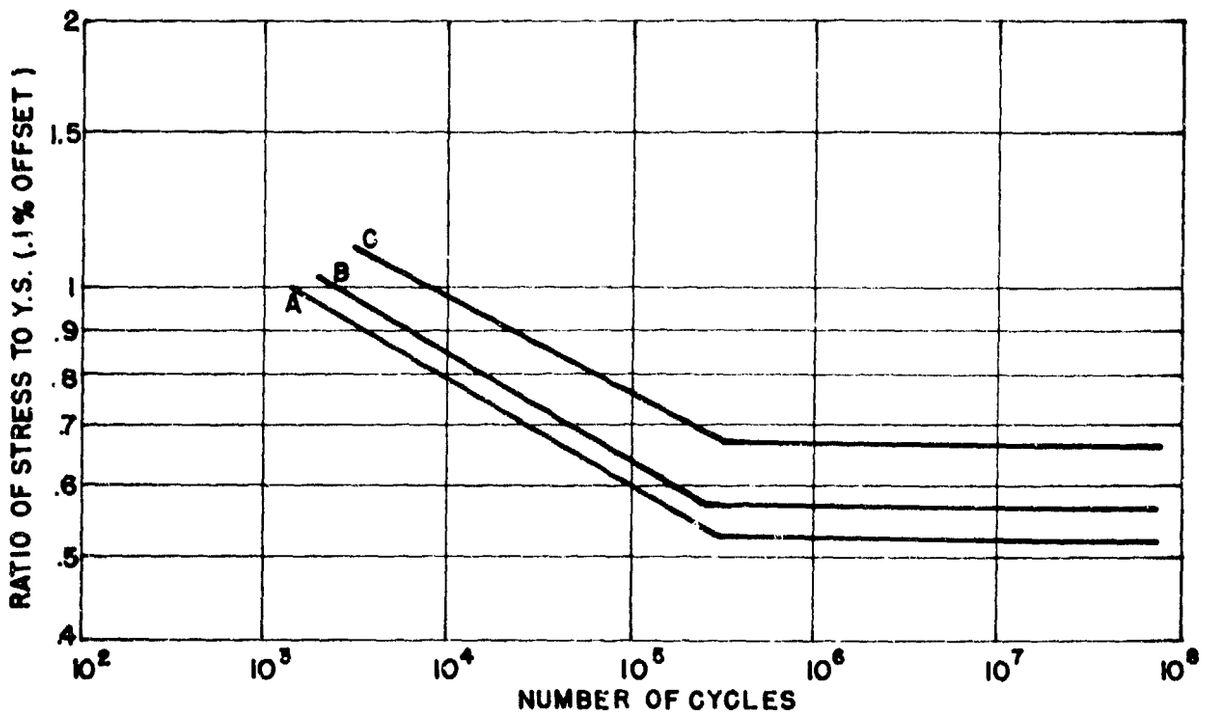


FIGURE 85

EFFECT OF STRENGTH LEVEL ON THE FATIGUE PROPERTIES OF A TEMPERED MARTENSITIC STEEL (YIELD STRENGTH A > B). (AFTER HOLLOMON ETAL[248])

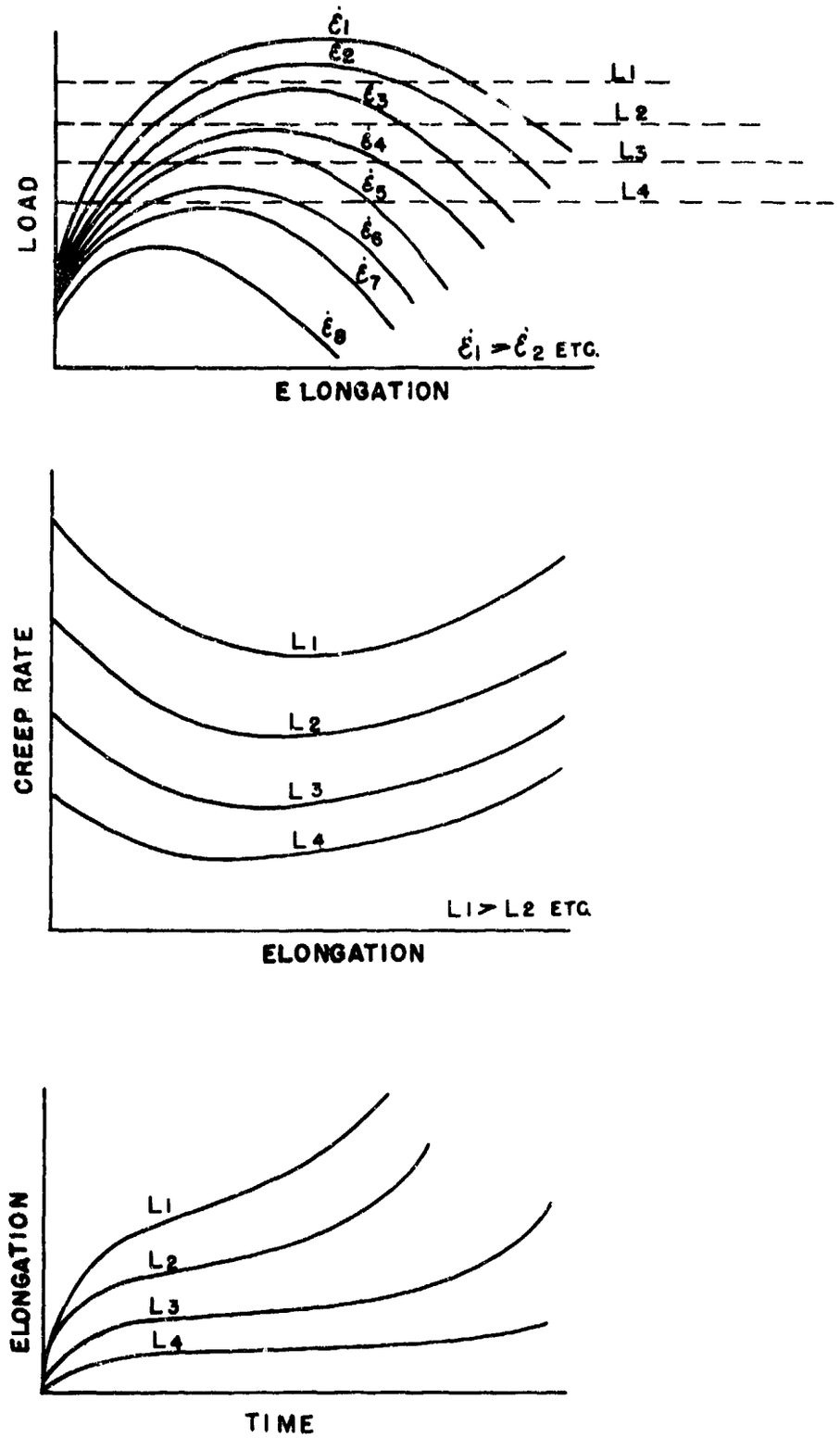


FIGURE 86

SCHEMATIC CONSTRUCTION OF CREEP CURVES FROM LOAD ELONGATION CURVES (AFTER ZENER & HOLLOMON[177])

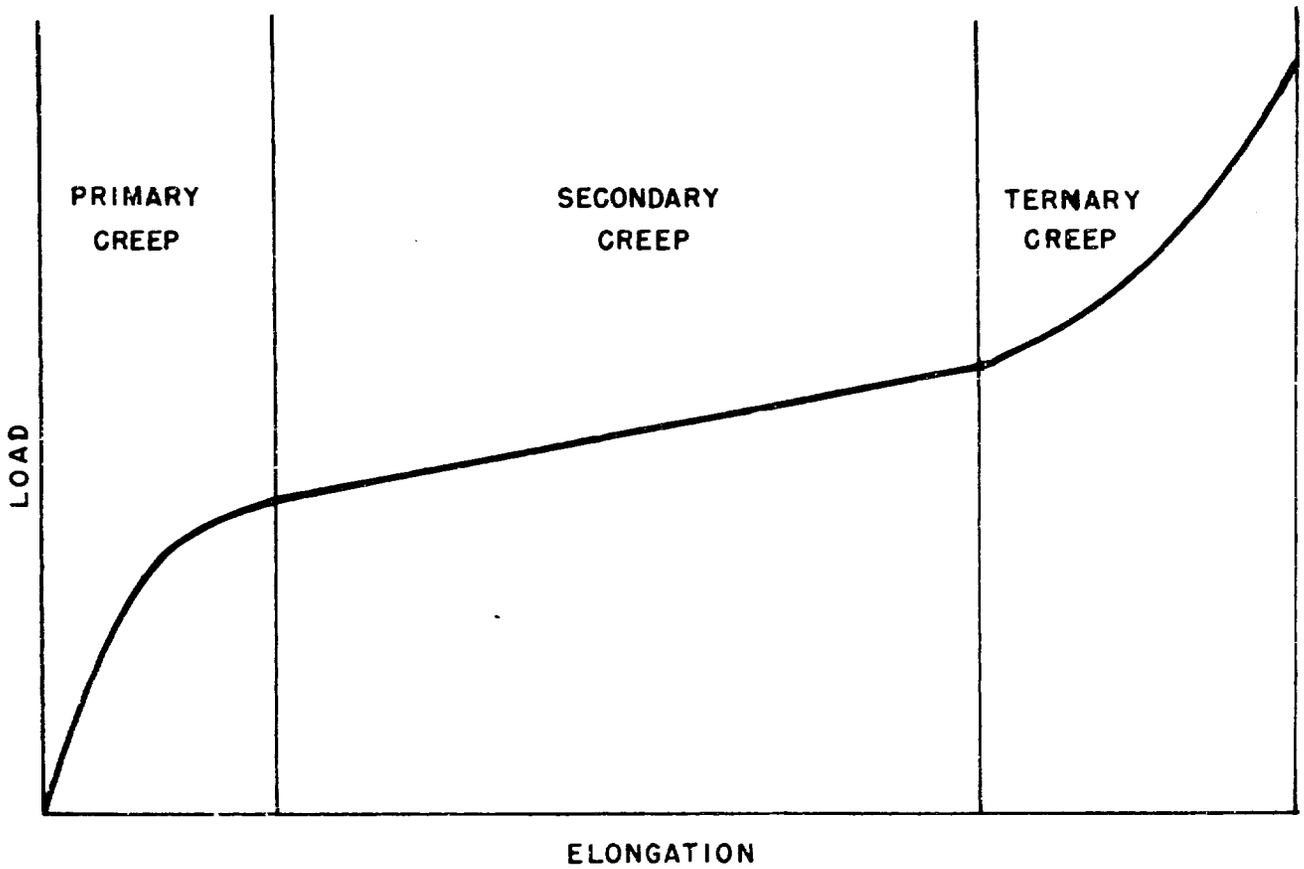


FIGURE 87

THREE STAGES OF CREEP

Suggested Reading

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V. QUENCHING

The need for cooling steel parts rapidly from the austenitizing temperature range through the pearlite and bainite ranges in order to form martensite was discussed in Chapter I. The fundamental factors governing heat flow in steel parts have been discussed in Chapter II. The application of the knowledge of heat flow to cooling steel parts rapidly from the austenitizing temperature through the pearlite and bainite ranges is the subject of the present chapter.

Practical Quenching Media

Quenching is ordinarily carried out in a fluid medium, commonly water, oil, or air. The temperature range of most rapid transformation from austenite to pearlite within a steel part generally corresponds to temperatures at which aqueous and oily media(132) provide stage A or stage B cooling. The temperature range of most rapid transformation from austenite to bainite generally corresponds to temperatures at which oily media provide stage B or C cooling while aqueous media provide stage A or B*. To minimize the possibility of the pearlite and bainite transformations, it is usually desired that the rate of heat flow across the steel-medium interface be high during the A and B stages. A high film coefficient and high quenching severity is therefore desired in these stages.

A steel-water interface has a very high film coefficient during the B stage(137). It is principally for this reason that water is so often used as a quenching medium. Unfortunately, the quenching severity of a water is relatively low during the A stage, perhaps 1/10 that in stage B(137). Moreover, the A stage tends to persist down to low temperatures. To obtain satisfactory quenches in aqueous media, measures are frequently employed that raise the temperature of transition from stage A to stage B. Usually the water is moved relative to the part surface. This may be done by moving the part, the water, or both. Agitation of the water provides movement past the steel surface, as well as turbulence. The turbulence tends to raise the transition temperature and also the quenching severity in both stages. The water may be impelled against the part by pressure, through a nozzle or as a spray. Such methods can be used to insure movement of the water through recesses in the part, where otherwise the A stage, with its slow cooling, would persist.

Approximate values for the quenching severity of various media are given in Table II. These values are for the surface temperature range

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*For high HD values the temperature difference between center and surface is such that the surface enters the C stage while the center is at a high temperature. However the rate of cooling of the center will continue to be controlled by the prior A or B stage cooling at the surface until the center itself reaches nearly the temperature at which the surface changed from B to C stage.

affecting the pearlite transformation within steel parts, and in the case of aqueous media are combinations of the values for A and for B stage cooling*. The values found range from 0.9 - 1.0 $inch^{-1}$ for still water to 4 $inch^{-1}$ for violently agitated water(65). The increase with degree of agitation is undoubtedly due primarily to an increase in the transition temperature and hence in the importance of the fast B stage relative to the slow A stage, rather than to changes of the quenching severity within either stage.

The transition temperature from A to B stage water-cooling can be raised by additions of sodium, calcium, or lithium chloride, sodium hydroxide, sodium carbonate, phosphoric acid, sulphuric acid (in small quantities) or other soluble inorganic compounds(132, 133, 137). As discussed in Chapter II, this effect may be associated with an increase of the surface tension between water and steam. Solutes, such as soap(132-134), that lower the surface tension, lower the transition temperature and are, therefore, generally to be avoided when hardening in water. The addition agent most commonly used is sodium chloride (common or rock salt), which is inexpensive and is not harmful to the operators. It does, however increase corrosion of the metal vessels that hold and circulate the water. Expense ordinarily prohibits use of any addition agent in those cases where the quenching medium is discarded after a single use. Where corrosion and expense considerations permit, sodium chloride brine rather than plain water can be used to raise the transition temperature and thus increase the rate of cooling in the pearlite and often in the bainite range. Quenching severity values for sodium chloride brine (probably 10% by weight sodium chloride) have been reported(65) to range from 2 $inch^{-1}$ for still brine to 5 $inch^{-1}$ for violently agitated brine (Table II). On the basis of surface tension data, it might be expected that calcium chloride would have a greater effect than sodium chloride in raising the transition temperature and hence the effective quenching severity of water.

Temperature of the water has a great effect upon the t mperature of transition from A to B stage cooling; the hotter the water the lower the transition temperature(132-134). Figure 88 provides a good example. An increase of water temperature decreases the cooling rate (quenching

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*These values (except those for salt) were determined not by heat flow measurements but by studies of the extent of pearlite transformation in steel cylinders of various sizes quenched in the media(254). Another method of obtaining an average H-value has been suggested: The extent of transformation in a steel when quenched in the unknown medium is compared with that of a standard shape (such as the Jominy bar described below) quenched in a standard medium(255). The H-values in each stage can be obtained by determining the slopes of the cooling curves when plotted as in Figure 21. The values for salt quenching given in Table II were found in this way. Another method is to supply heat continuously to a part while in the quenching medium, and to measure the surface temperature and the heat input when the surface temperature becomes constant.

severity) within each stage only slightly. However, for the results illustrated in Figure 88, raising the water temperature from 70°F. to 122°F. (21°C. to 50°C.) lowered the transition between A and B stages from about 1580°F. (860°C.) to about 1080°F. (580°C.). Increasing the water temperature to 165°F. (74°C.) lowered the transition to about 670°F. (350°C.). To obtain a high cooling rate in the pearlite and bainite temperature ranges, it is, therefore, important to keep the water cold. It has been reported that satisfactory cooling rates in these ranges can be obtained with warm brines(256); presumably the presence of the salt keeps the transition temperature moderately high despite the medium being warm.

In addition to the problem involved in maintaining a high transition temperature from stage A to stage B with aqueous media, difficulties arise from the relatively high quenching severity (cooling rate) of these media during the C stage. The temperature range of the transformation of austenite to martensite falls within the C stage(137). To minimize the likelihood of quench-cracking, it is usually desirable to decrease the temperature gradient throughout the part by cooling slowly through this temperature range. A low quenching severity in the C stage is, therefore, advantageous in cases, such as with aqueous media, where this stage commences below the range of rapid transformation to bainite. Additions of various solutes to water seem to have little effect upon its quenching severity in stage C (132, 133). Motion and turbulence increase the severity somewhat. Changes in medium temperature seem to have little effect upon the film coefficient or quenching severity(133). However, an increase in medium temperature does decrease the temperature difference across the interface and so tends to decrease the rate of heat flow. This effect is appreciable only at temperatures not far above that of the medium; nevertheless, since it is at such temperatures that quench-cracking is most likely to occur, the effect may be of some aid in avoiding cracking. In general, little advantage can be taken of this with aqueous media, since raising the medium temperature has such a strong tendency to lower the temperature of transition from A to B stage.

Oil is used as a quenching medium primarily because it has a much lower quenching severity in stage C than water(137), and is, therefore, generally much less likely to crack the piece. Because quenching oil has a higher boiling point than water, the C stage commences at a higher temperature. This is advantageous in providing protection against cracking but has the disadvantage that the C stage may commence before the center of the part has passed the temperature of the bainite nose, thus increasing the likelihood of bainite formation.

The problem of raising the temperature of transition from the A stage to the fast B stage occurs with oil as well as with water. However, it is not as important in the case of oil, because the difference in cooling rate between the two stages is not as great. Indeed, the quenching severity during stage A is, for some oils at least, greater than for water (137). Motion and turbulence tend to raise the transition temperature of oil, and certain addition agents probably also do so. Changing the temperature of oil has much less effect than changing the temperature

of water. Presumably this is because the direct effect of increasing medium temperature in maintaining a vapor layer (stage A) is counteracted by the decrease in viscosity of the oil and corresponding increase in its turbulence.

The quenching severity (film coefficient) during the third stage is probably not decreased by raising the oil temperature; more likely it is increased somewhat. However, the oil can be heated sufficiently to decrease the temperature difference across the interface significantly in the martensite temperature range, and so to decrease the rate of heat flow in the range where this is desirable. In practice, quenching oil is usually used somewhat above room temperature; temperatures almost up to the flash point are sometimes employed.

The quenching characteristics of various oils differ appreciably (133), but the effects of commonly determined characteristics such as composition and viscosity upon the quenching characteristics are not yet completely understood. Little is known even of the differences between the behavior of petroleum oils and of fatty oils. It does appear to be generally true that as the boiling point is raised, the temperatures of transformation from stage A to B stage and from B to C also increase. Quenching severity values ranging from .25 - .30 $inch^{-1}$ for still oil to .8 - 1.1 $inch^{-1}$ for violently agitated oil have been reported(65). (Table II). In selecting a quenching oil not only the quenching characteristics but also others, such as flash and fire points, stability in service, and cost, are usually of interest.

The use of emulsions of oil in water has been suggested to achieve cooling rates intermediate between oil and water. Such emulsions have, however, the high quenching severity of water during stage C as well as low temperatures of transition from stage A to B (135). They, therefore, appear to offer no advantage over water.

The use of molten salts as quenching media has been increasing. Eutectic mixtures of alkali nitrates and nitrites are most commonly used because of their low melting points. The advantage of salt quenches is that the medium can be made hot enough to greatly decrease the temperature difference across the metal-medium interface, in the martensite temperature range. This slows the rate of heat flow and decreases the temperature gradients within the part, in the range where cracking is most likely to occur. The salt temperature can be made higher than the M_s temperature so that the temperature within the part can be completely equalized before the steel is cooled through the martensite range. Moreover, this subsequent cooling (for which, ordinarily, the part must be removed from the salt) can often be carried out slowly, as in air. Many of the salts used will boil or decompose to gaseous products at the temperatures used for austenitizing, so that B stage and perhaps A stage cooling can occur. It appears, however, that salt quenching may sometimes be entirely C stage. Because A stage cooling is less prominent, agitation probably has less effect on salt quenches than on oil, though it undoubtedly is important. Salt temperature, also, probably does not affect the quenching severities very greatly, but

does, of course, appreciably affect the temperature difference across the interface at the lower surface temperatures. The quenching severities of salt quenches (at surface temperatures corresponding to metal temperatures in the pearlite range) are of the same order of magnitude as oil quenches, tending perhaps to be somewhat lower (Table II).

Quenching in molten metal has the same advantage of a high medium temperature as salt quenching. The quenching characteristics of the different metals and alloys do not appear to have been systematically investigated. Lead is commonly used in production heat-treatment. Lower melting (but more expensive) eutectic mixtures, such as Wood's metal, have been used in laboratory work. As the metals and alloys used generally do not boil or decompose at the austenitizing temperatures, they provide quenching that is entirely C stage. Lead has been stated to give quenching severities of the same order as oil and salt(257, 258); it may give somewhat higher severities.

Air quenches have often been used to minimize gradients within a part by providing a low quenching severity and hence a low cooling rate in the martensite range. They accomplish this purpose, but at the same time provide a low film coefficient in the pearlite and bainite temperature ranges. As a result, only steels that transform very slowly in the pearlite and bainite ranges will harden to martensite upon air-cooling (except in very thin sections). Air-cooling is wholly C stage, and the quenching severity (film coefficient) does not change much with the surface temperature or with the medium temperature. Heat transfer by radiation between the part and the surrounding walls often plays a considerable part in so-called "air-cooling", so the walls as well as the air must be considered. The quenching severity of still air has been reported(254) to be of the order of .02 $inch^{-1}$. (Table II.)

Gaseous atmospheres that do not decarburize the steel surface are sometimes used for cooling in place of air. Hydrogen and helium have also been used, in laboratory work, to provide quenches more rapid than air, and carbon dioxide to provide quenches slower than air. For equal gas velocities, pressures, and degrees of agitation, the quenching severity (film coefficient) appears to increase with the thermal conductivity and presumably with the heat capacity per unit volume of the gas.

Sometimes pieces are quenched by placing them between cold steel dies. Such cooling may be considered to be C stage; the quenching severity does not change much with temperature. (It must be remembered that not only the quenching severity or film coefficient but also the thermal properties of the die metal itself are of importance if an appreciable amount of heat from the part reaches the back or outside surface of the dies.) It does not appear that quenching severities as high as those attained with oil quenching can be obtained by using steel as the quenching medium(133). This type of die-quenching does have the advantage of holding the part to a given shape during cooling, which appears to be of aid in minimizing distortion.

Equivalence of Cooling Curves

Criteria of Equivalence

For design purposes it is desirable to determine whether or not the cooling curve at one position in a part is metallurgically equivalent to the cooling curve at some position in another part. The curves will never match exactly in all particulars if the two cases differ in any detail. Even for homogeneous austenite of the same composition and grain size quenched from the same temperature into media at the same temperature and having quenching severities independent of surface temperature, the cooling curves will be of different shapes if the shapes or sizes of the parts, the positions within the parts, or the quenching severities differ in the two cases. To talk profitably of equivalence of cooling curves that have different shapes it is necessary to establish a criterion defining the cooling. For studies of hardening this criterion should be based upon consideration of those characteristics of the cooling curve that determine whether the steel will transform in the pearlite or bainite ranges or will remain untransformed down to the martensite range.

Among the criteria that have been suggested are the cooling rate at a given temperature(259), such as 1300°F. or 1000°F., and the time to cool through a given temperature range(260, 261), such as 1100°F. to 900°F. or 1350°F. to 900°F. The half-temperature time(262), that is, the time to cool from the initial temperature to one midway between the initial and medium temperatures, has also been proposed. Another criterion that has been suggested is the cooling coefficient(263), which is defined as the constant ratio, between the (fractional) cooling rate and the fractional temperature, that is approached as the fractional temperature decreases (c of equations (2-1) and (2-3). The slope of the lower portions of the cooling curves of Figure 21 is a measure of the cooling coefficient.

If the HD product is low and constant, the cooling curve for any point within the part is not far from exponential. Two curves for which the HD product is constant and low (not over 0.1) will, therefore, match approximately if any of the suggested criteria are equal. Under such conditions all of the criteria will be equally valid. This will be true whether or not the two initial temperatures are equal. It will also hold approximately even if the two media temperatures differ somewhat (except at temperatures approaching or below that of the warmer medium). The cooling curves will not match, however, if the parts are made of steels having thermal diffusivities that change appreciably with temperature within the range of cooling (unless the two diffusivities change in the same way).

It has been pointed out in Chapter II that the cooling curve at any point in a quenched part approaches an exponential curve as the temperature of the part approaches that of the medium, if the thermal characteristics of the part and of the part-medium interface do not change greatly with temperature. The fractional temperature at which the cooling becomes approximately exponential is shown in Figures 89 and 90 for the center and surface of cylinders and plates. The derivation from exponential cooling

is small for temperatures somewhat higher than those indicated in the figures. Only that part of the cooling curve below the range of austenite stability significantly affects the decomposition of austenite. For a part quenched from temperatures reasonably far above the A_3 and A_{cm} temperatures into a medium not much over room temperature, the cooling of the center does not differ much from exponential through and below the temperature range where transformations can take place (provided the thermal diffusivity and quenching severity do not vary much with temperature). This will be true even though the center does not cool exponentially above the transformation range if the HD is high (Figure 21). At the surface of plates and cylinders, however, the cooling within the transformation range differs considerably from exponential for HD values much over 0.5. For positions half way between center and surface the cooling is exponential within about the same limits as at the center.

For those cases where the cooling is approximately exponential over the range of metallurgical interest, any criterion, based upon the exponential relation, that does not involve cooling at higher temperatures will be valid. Thus, cooling rate at any temperature within the range of exponential cooling, time to cool over any temperature interval within this range, or cooling coefficient can be used. Criteria involving cooling at the upper end of the range of exponential cooling (such as cooling rate at 1300°F, or time to cool from 1350°F. to 900°F.), will involve more error than criteria based upon lower-temperature portions of the cooling curve. A criterion (such as half-temperature time) based to a large extent upon portions of the cooling curve above the range of exponential cooling would be expected to involve considerable error.

The cooling curves at and near the surface of parts for high HD values are not close to exponential within the range affecting the pearlite or even the bainite transformation. To properly compare such curves, it would be necessary to integrate over them the amount of transformation occurring for the steel under consideration (or the fraction of the time before transformation begins) and compare the resulting integrated values. To do this it would be necessary to know the transformation behavior of the steel as a function of time at all temperatures, as well as the quantitative effect of holding at one temperature within the transformation range upon subsequent transformation at a lower temperature. (This last is, as has been discussed in Chapter I, fairly well established for the pearlite transformation but not for the bainite.) The equivalence of the cooling curves would then depend on the transformation characteristics of the steel. Two cooling curves equivalent for one steel might not be equivalent for another. Practically no work has been done toward solving this problem.

The procedure generally adopted for comparing cooling curves for positions at or near the surface for high HD values has been to use some arbitrary criterion, without considering its justification. Cooling rate at a selected temperature, time to cool through a selected temperature interval, and half-temperature time have all been used. They give different results. Moreover, with the first two criteria, the result (presence or absence of equivalence) will depend upon the particular temperature or temperature range used in the criterion. Cooling coefficient has not been

used for comparisons under the condition mentioned; it would not be expected to be of much value here, since the coefficient is the same for all positions in a part. No systematic investigation of the merits of the various empirical criteria for cooling conditions at or near the surface during high HD quenches has been made; one investigation(262) indicated that for one steel (SAE 3140) half-temperature time gave a better indication of the extent of transformation than did cooling rate at 1000°F.

Equivalence of Center Cooling Curves

For obtaining full hardening throughout a part the cooling curve at the center is of primary interest. For comparisons of such curves, the cooling coefficient seems to be at least as good a criterion as any that has been proposed, and has the advantage of being simplest to compute from the laws of heat flow. To avoid the confusion that would be involved in comparing each of the center cooling curves for a number of sizes, shapes, and severities of quench directly to each of the others, the cooling of a standard shape when quenched into a medium of specified severity can be used as a standard with which to compare all the other curves. The "ideal round" has been widely adopted as the standard(262). This is a round bar of infinite length (right circular cylinder) having a constant thermal diffusivity and quenched in a medium providing an infinite film coefficient (infinite severity of quench H and so infinite value of the product HD). An infinite film coefficient means that the metal surface is instantaneously cooled to the medium temperature, providing the fastest possible quench. Figures 91 to 96 show the size of ideal round having the same cooling coefficient as various sizes of rounds (solid cylinders), plates, and hollow (right circular) cylinders, quenched in media providing severities of 1, .3, and .02 $inch^{-1}$ (Table II). These severities were chosen as the estimated severities obtained in quenching medium-alloy steels from the austenitizing range into still water, still oil, and still air, at room temperature (Table II)*.

For squares, the equivalent ideal round size is 100-110 per cent of that for rounds of the same thickness. For cubes, the ideal round size is 83-91 per cent that for rounds of the same thickness, and for spheres, 80-91 per cent. The effect of cooling the ends of round and square bars may be neglected unless the length is almost as small as the thickness(144), in which case the values for cubes or spheres may be used. Rectangular bars are intermediate between square bars and plates. Most parts can be treated as approximations to the simple shapes mentioned, if it is remembered that the cooling at the center (if H is approximately constant over the surface) is primarily a function of the smallest dimension (thickness), the other dimensions having relatively little influence.

A special shape that is of considerable interest in metallurgical design for hardening is the end-quench hardenability or Jominy bar(264).

*Equivalence charts covering all quenching severities are available in the literature(263).

This is a cylindrical piece 1 inch in diameter and 3 or 4 inches long, having ends cut square. It is cooled with its axis vertical and a stream of water applied to the bottom end while the sides and top cool in still air. The water is at a temperature of 75 ± 5°F. (24 ± 3°C.) and passes through a 1/2 inch diameter opening 1/2 inch below the bottom of the piece. The flow of water is such that with the specimen removed the water column rises to a height of 2-1/2 inches above the opening. The cooling conditions within the Jominy bar during this quench vary little across the diameter, but vary considerably along the length of the bar. Neglecting the air-cooled end, cooling of the Jominy bar may be considered as a combination of water-quenching half of a six or eight inch plate and air-cooling a one-inch round. It has been reported(265) that the water-quenching provides a quenching severity of about 2.33 $inch^{-1}$, and the air-cooling about .022 $inch^{-1}$. It appears to make little difference whether the plate thickness is considered to be 6 or 8 inches, corresponding respectively to a 3 and a 4 inch long bar.

Since the cooling conditions at least near the water quenched end of the Jominy bar are similar to those near the surface of a quenched plate of high HD value (14 to 19), no theoretical basis for comparison with cooling conditions at the center of ideal rounds or other shapes exists. It has been reported that reasonably good results are obtained by using the empirical criterion of half-temperature time, and a curve supported by some experimental data has been published for comparison of distances along the Jominy bar with sizes of ideal rounds having the same half-temperature time at the center(265). This curve has been replotted on logarithmic scales (on which it is linear over most of the length) and is shown in this form in Figure 97*. By combining this curve with those of Figures 91 to 96, Figures 98 to 103 have been derived. These figures permit the direct determination of the position on the Jominy bar having cooling conditions equivalent to those at the center of rounds, plates, and hollow cylinders under various severities of quench.

In using Figures 98 to 103, it should be kept in mind that the comparisons are for shapes, ideal rounds, and Jominy bars having equal and constant thermal diffusivities, that the severities of quench are considered constant over the temperature range of the cooling, and that the cooling media are taken to be at room temperature. The thermal diffusivity of austenite is not independent of temperature and varies somewhat with composition, decreasing as its carbon and alloy contents increase**. The transformations that occur introduce heat evolutions which effectively change the diffusivity, and the diffusivity of the decomposition products differs from that of austenite. The severities of water and oil cooling change markedly during the quench (because of the three-stage nature of such

*The portion of the curve at and beyond 2 inches from the quenched end was calculated by the authors and includes the effect of air-cooling the top end of the bar.

**The value 0.009 $in.^2$/sec. for the diffusivity of austenite has been suggested as a first approximation for steels of moderate carbon and alloy content(266).

quenches), and the severity (averaged over the appropriate temperature interval) is greatly influenced by such factors as the degree of agitation. The cooling medium temperature may differ somewhat from room temperature. Moreover, the criteria of equivalence used (particularly for cases involving the Jominy bar) are not exact. Hence the figures cannot be considered to be of high accuracy, and are intended simply to be used as first approximations in design work. With this in mind, they have (by using H values for still rather than agitated media) deliberately been made conservative, in the sense that the equivalent ideal round sizes and Jominy distances indicated are somewhat lower than those that have generally been found experimentally for the types of media indicated.

TABLE II

APPROXIMATE QUENCHING SEVERITY (H) OF VARIOUS MEDIA

(IN PEARLITE TEMPERATURE RANGE)

Circulation or Agitation	Value of H in $Inch^{-1}$ For				
	Brine*	Water*	Oil*	Salt	Air**
None	2	.9 to 1.0	.25 to .30	.20 to .25	.02
Mild	2 to 2.2	1.0 to 1.1	.30 to .35	----	----
Moderate	----	1.2 to 1.3	.35 to .40	.25 to .30	----
Good	----	1.4 to 1.5	.4 to .5	----	----
Strong	----	1.6 to 2.0	.5 to .8	----	----
Violent	5	4	.8 to 1.1	.4 to .5	----

*After Grossmann (65).
**After Grossmann and Asimow (254).

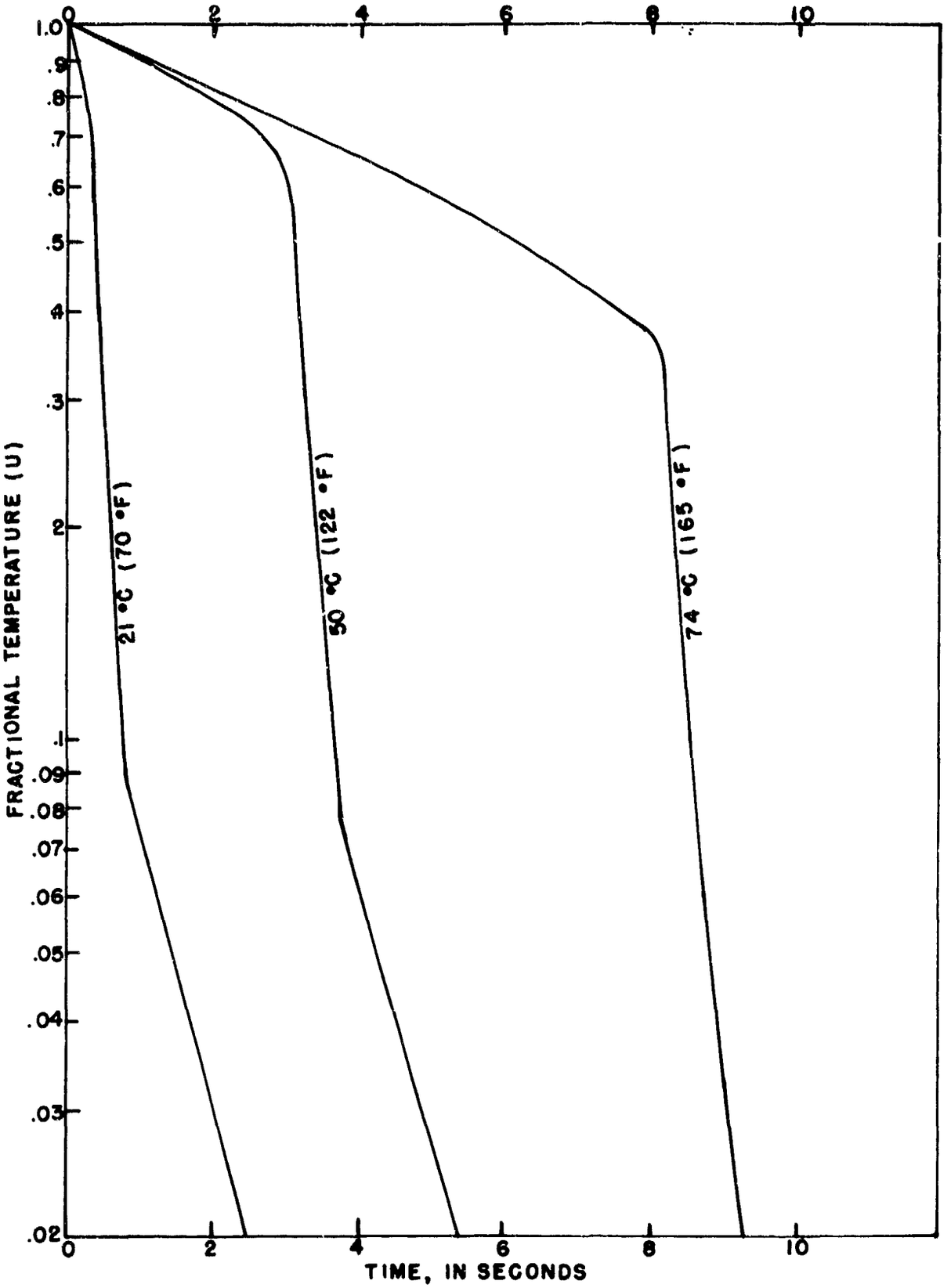

FIGURE 88

EFFECT OF WATER TEMPERATURE UPON COOLING CURVES

4 MM CHROMIUM-NICKEL SPHERE. TIMES MEASURED FROM 860 °C (1580 °F)
(FROM DATA OF WEVER[155])

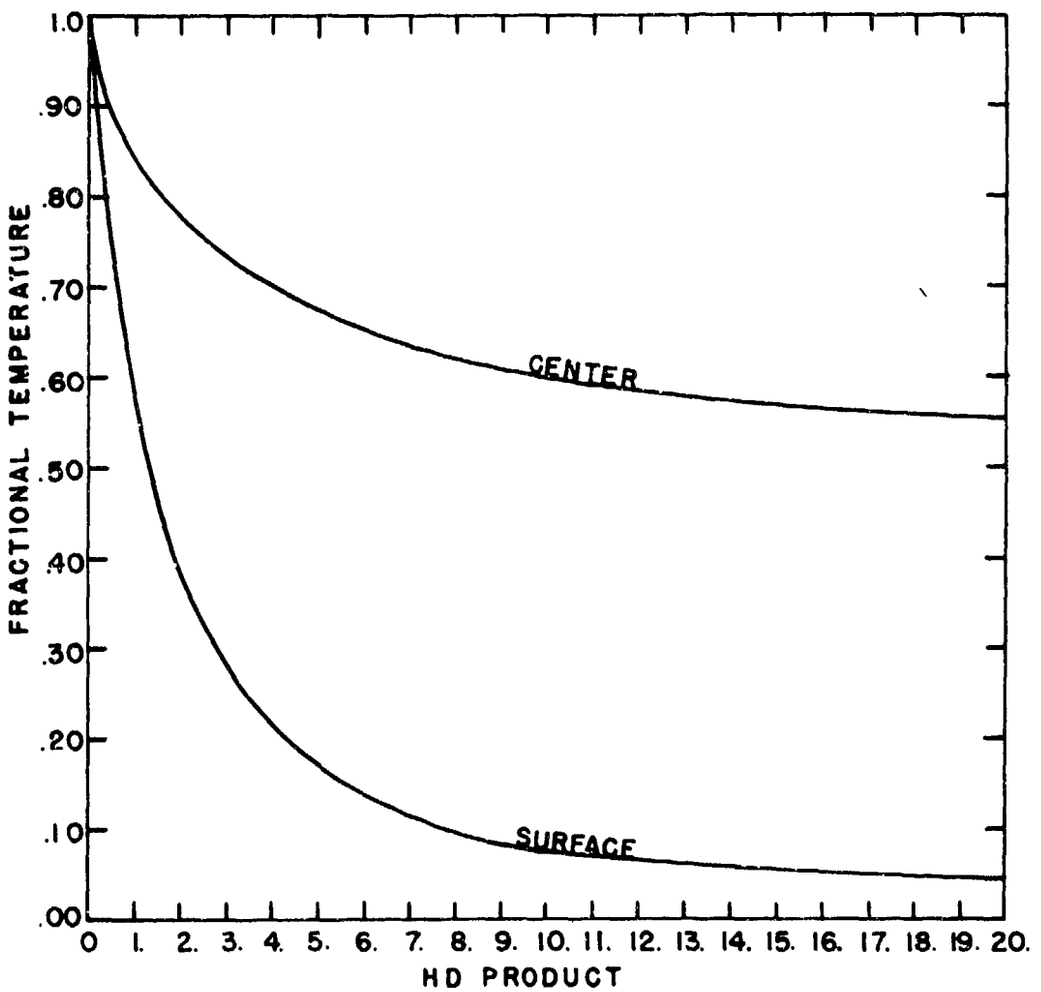

FIGURE 89

FRACTIONAL TEMPERATURE BELOW WHICH COOLING OF CYLINDERS IS APPROXIMATELY EXPONENTIAL.

(ESTIMATED FROM TABLES OF RUSSELL[141])

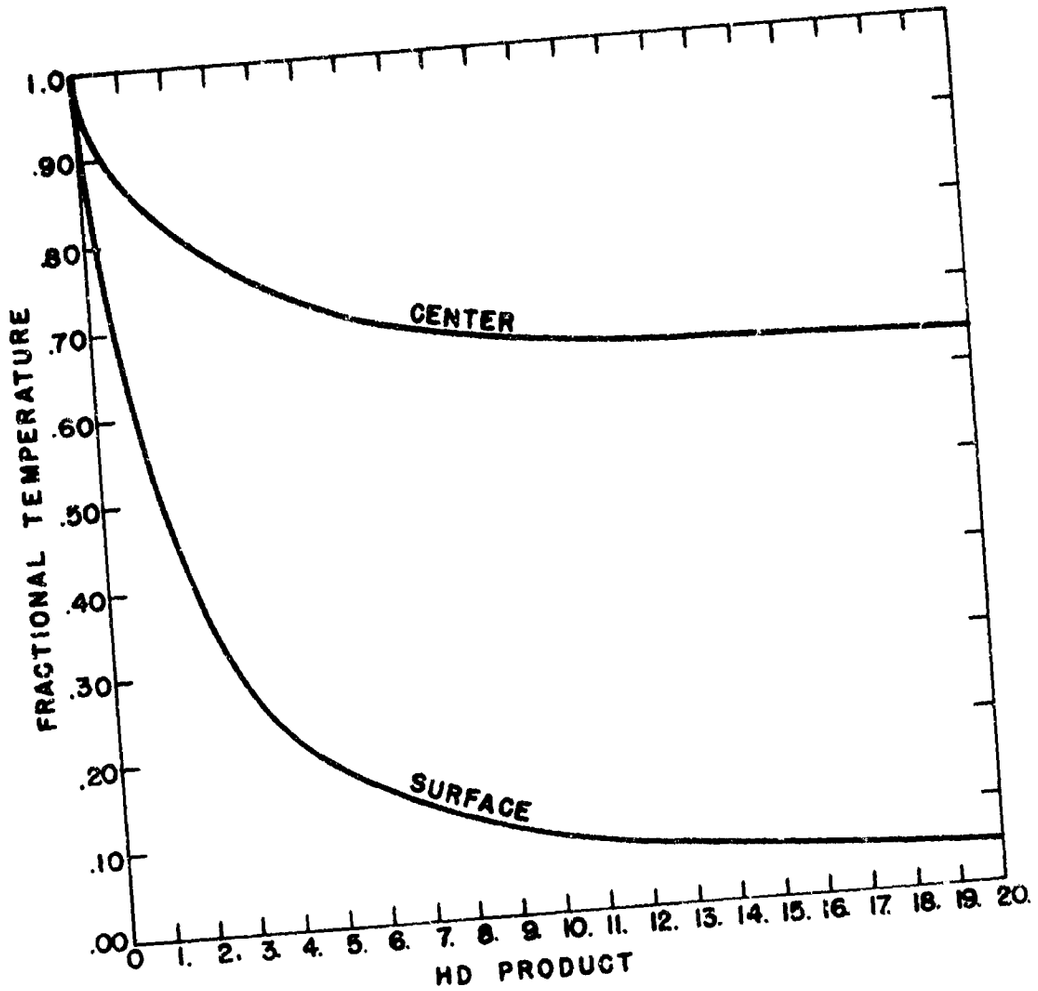

FIGURE 90

FRACTIONAL TEMPERATURE BELOW WHICH COOLING OF PLATES IS APPROXIMATELY EXPONENTIAL. (ESTIMATED FROM TABLES OF RUSSELL[141]).

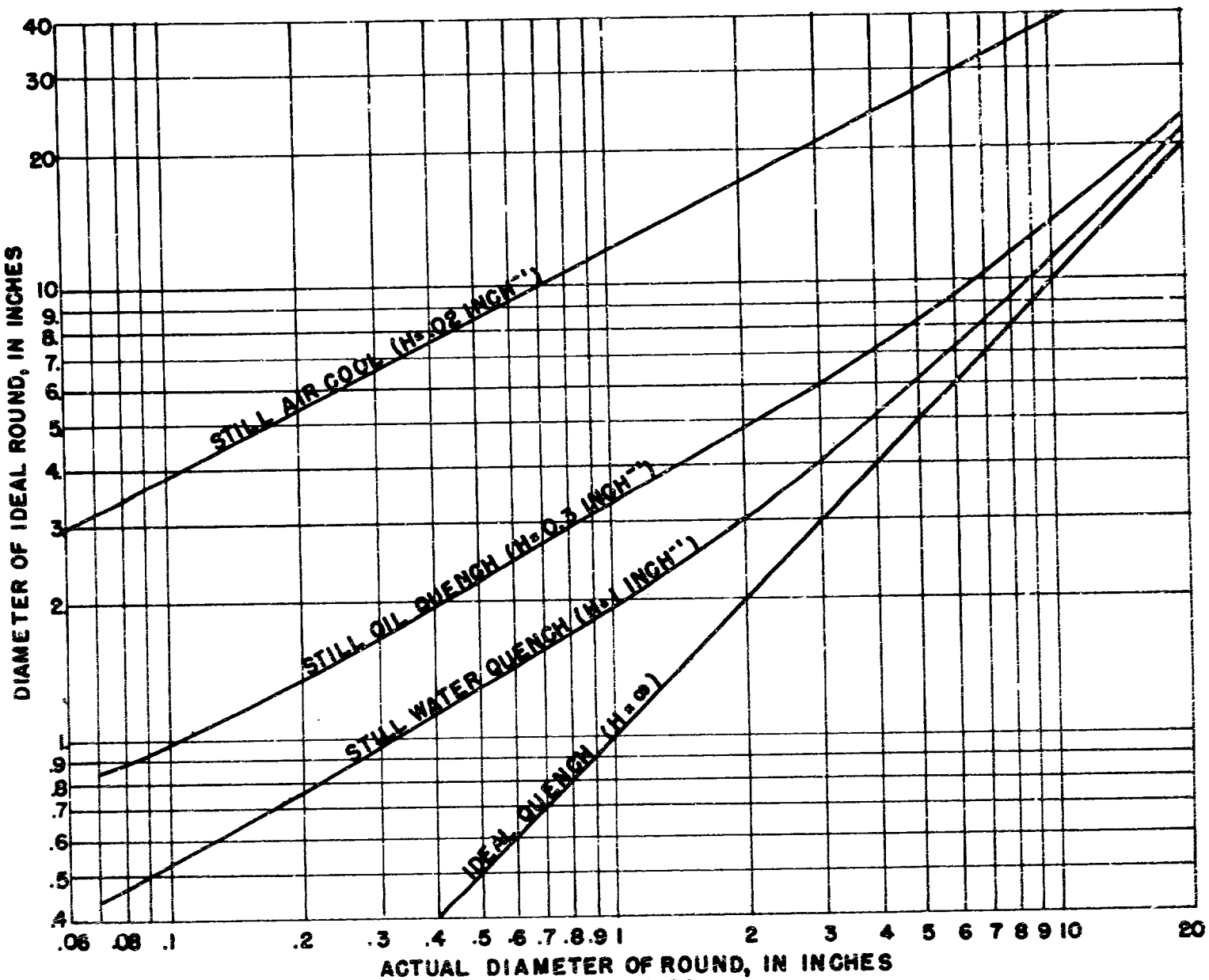

FIGURE 91

SIZES OF ACTUAL & IDEAL ROUNDS HAVING EQUIVALENT COOLING CONDITIONS

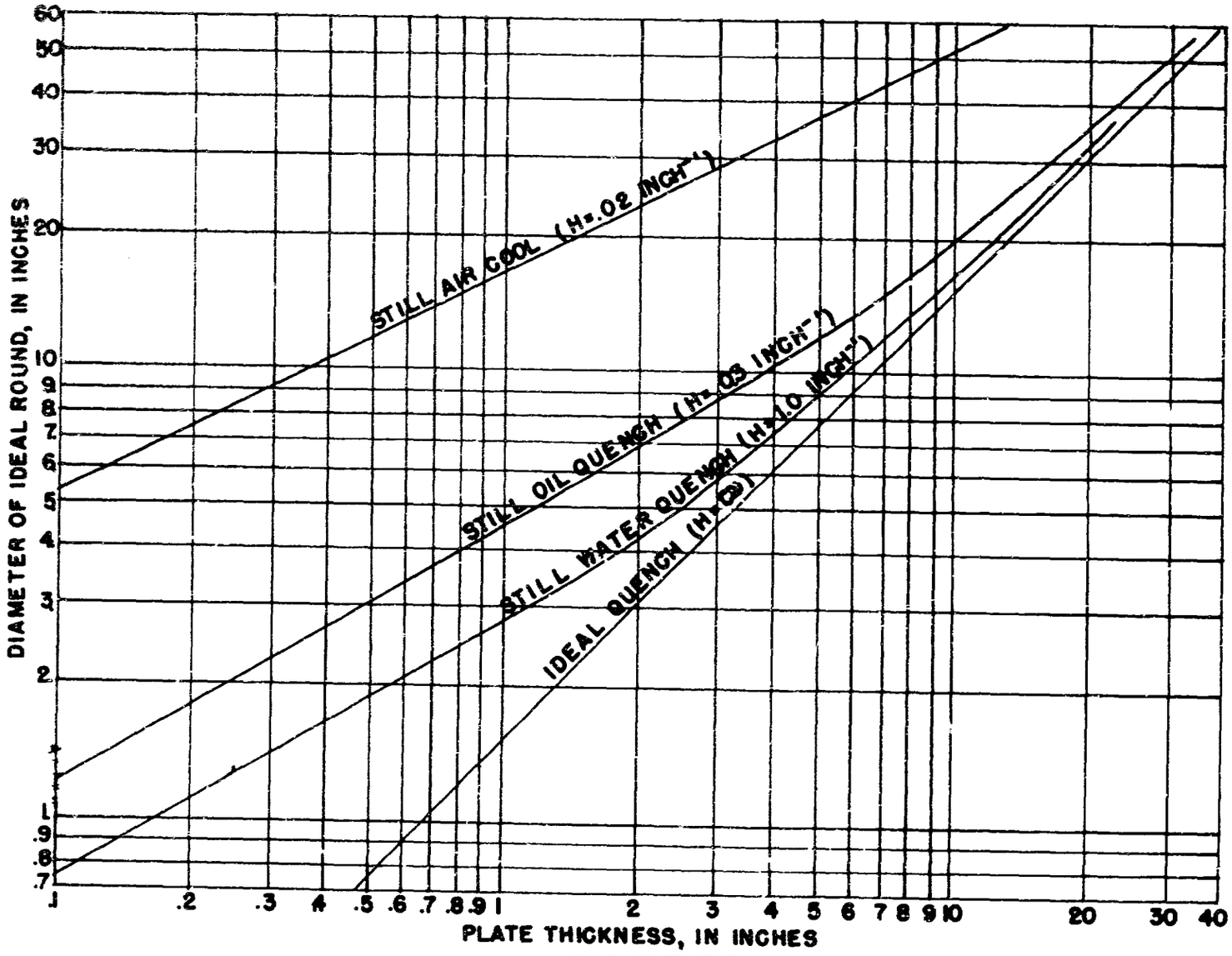

FIGURE 92

SIZES OF PLATE & IDEAL ROUND HAVING EQUIVALENT COOLING CONDITIONS

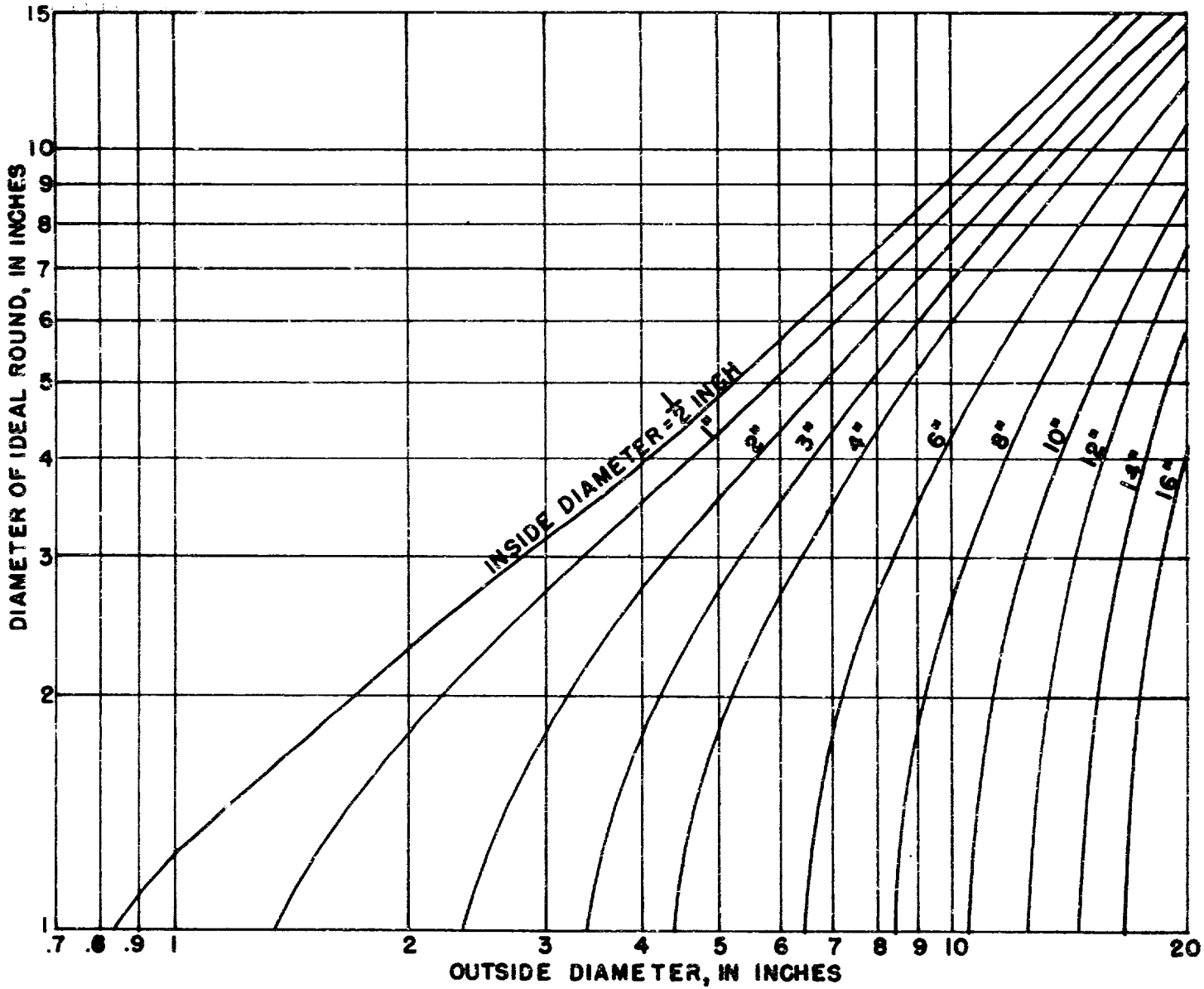

FIGURE 93

SIZES OF HOLLOW CYLINDER & IDEAL ROUND HAVING EQUIVALENT COOLING CONDITIONS. CYLINDER WATER QUENCHED INSIDE & OUTSIDE ($H = 1$ INCH^{-1})

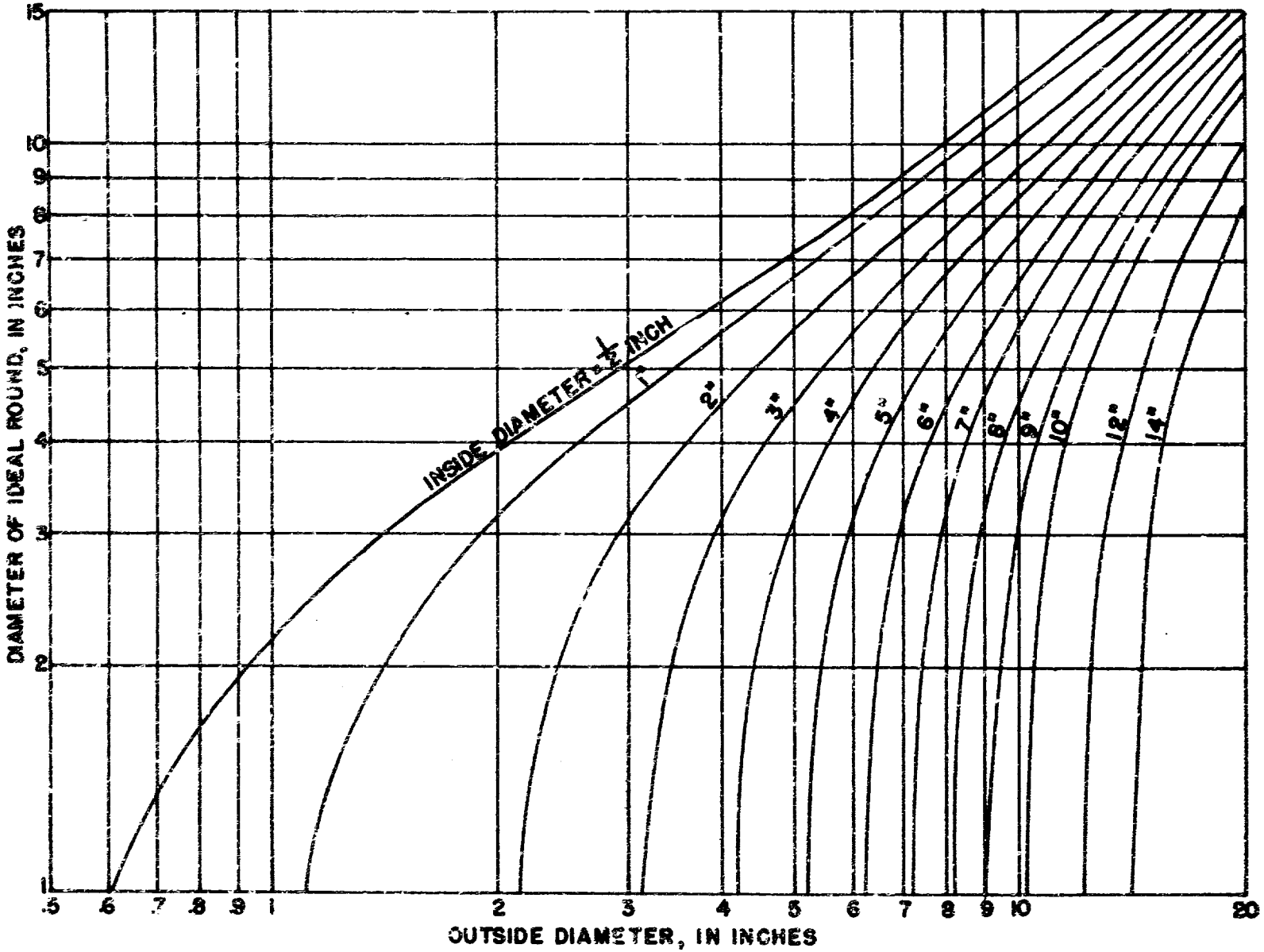

FIGURE 94

SIZES OF HOLLOW CYLINDER & IDEAL ROUND HAVING EQUIVALENT COOLING CONDITIONS
CYLINDER OIL QUENCHED INSIDE & OUTSIDE ($H = 0.3$ INCH^{-1})

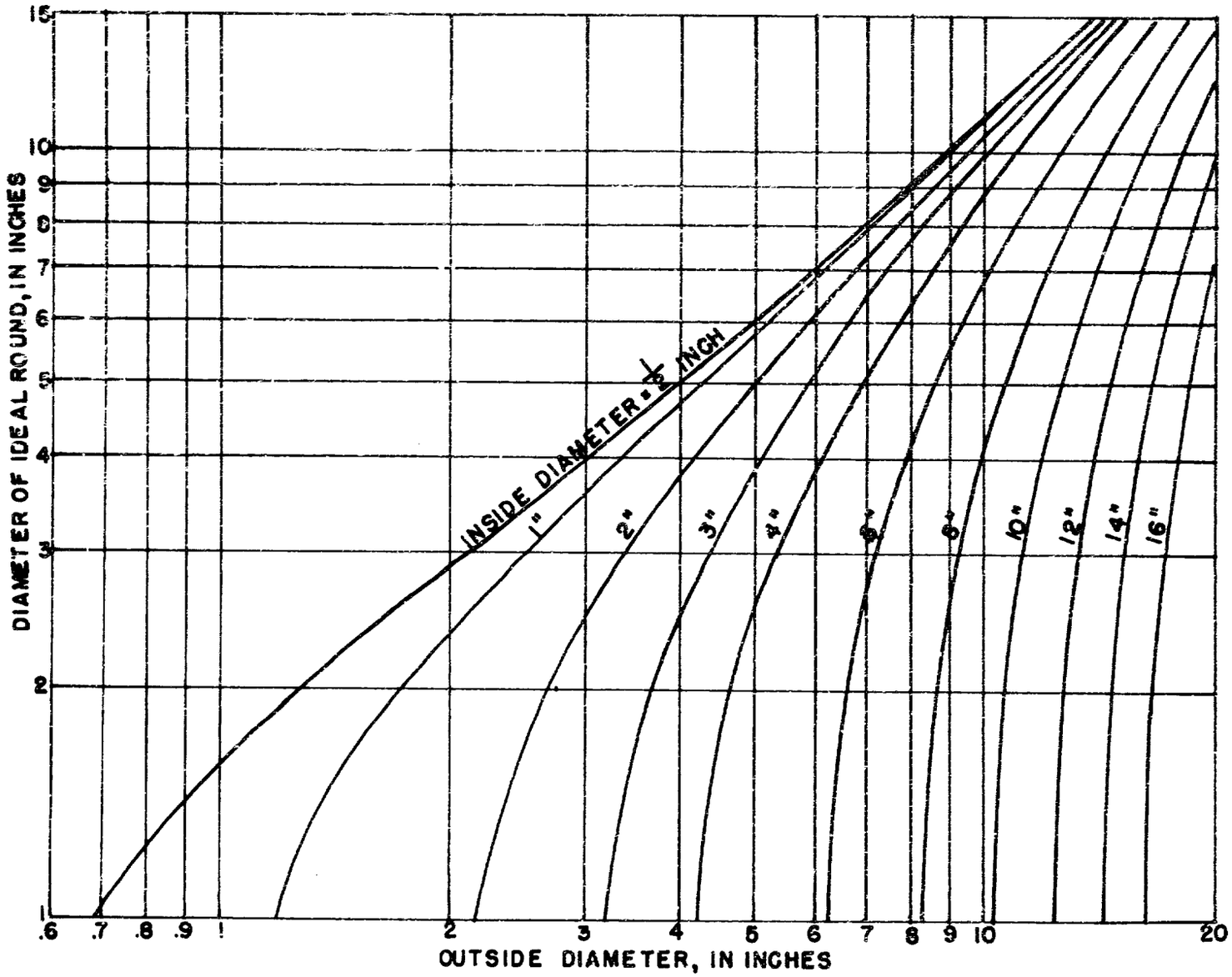

FIGURE 95

SIZES OF HOLLOW CYLINDER & IDEAL ROUND HAVING EQUIVALENT COOLING CONDITIONS
CYLINDER WATER QUENCHED, OUTSIDE ONLY ($H = 1\ \text{INCH}^{-1}$)

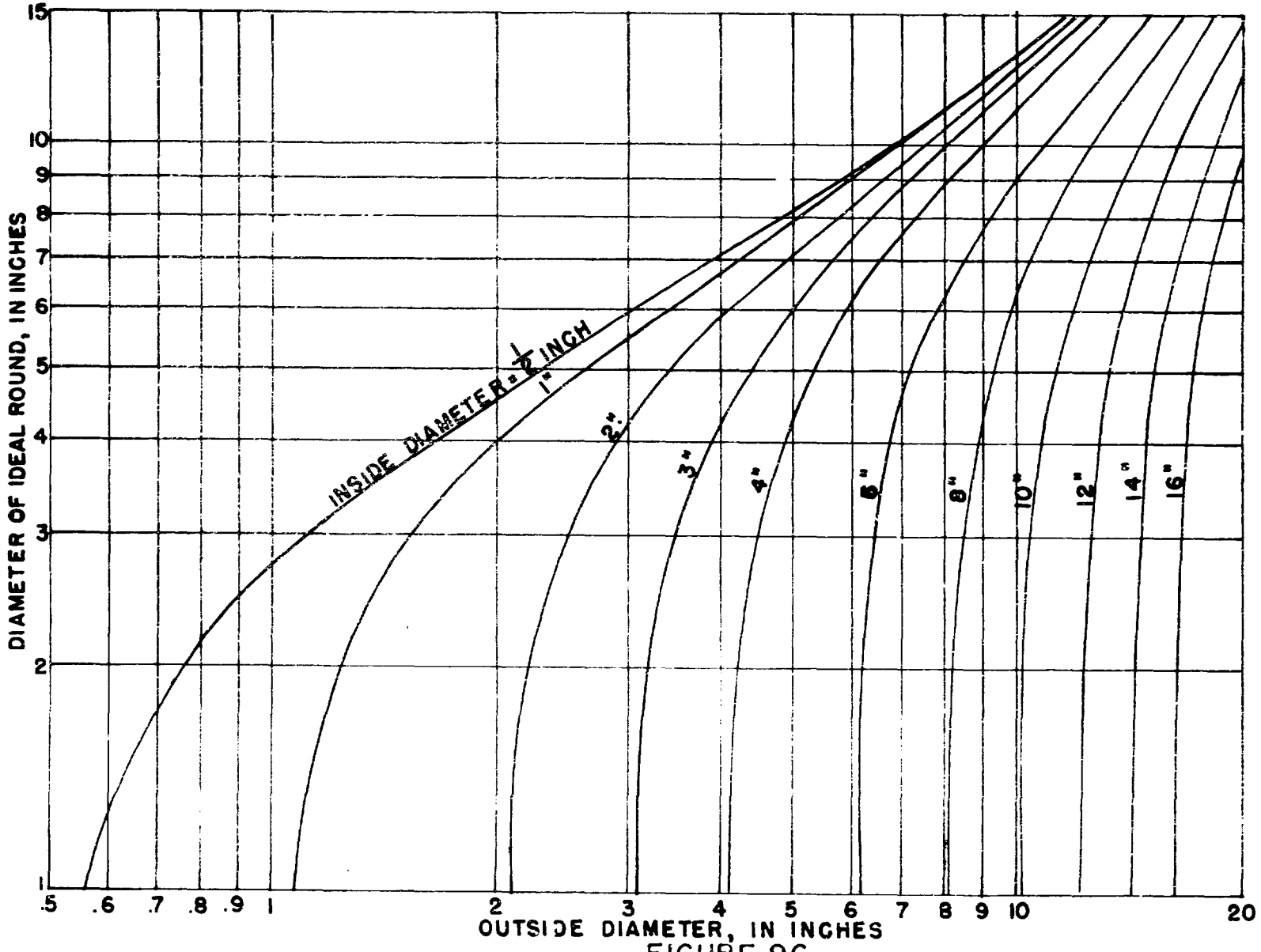

FIGURE 96

SIZES OF HOLLOW CYLINDER & IDEAL ROUND HAVING EQUIVALENT COOLING CONDITIONS
CYLINDER OIL QUENCHED, OUTSIDE ONLY ($H = 0.3$ INCH^{-1})

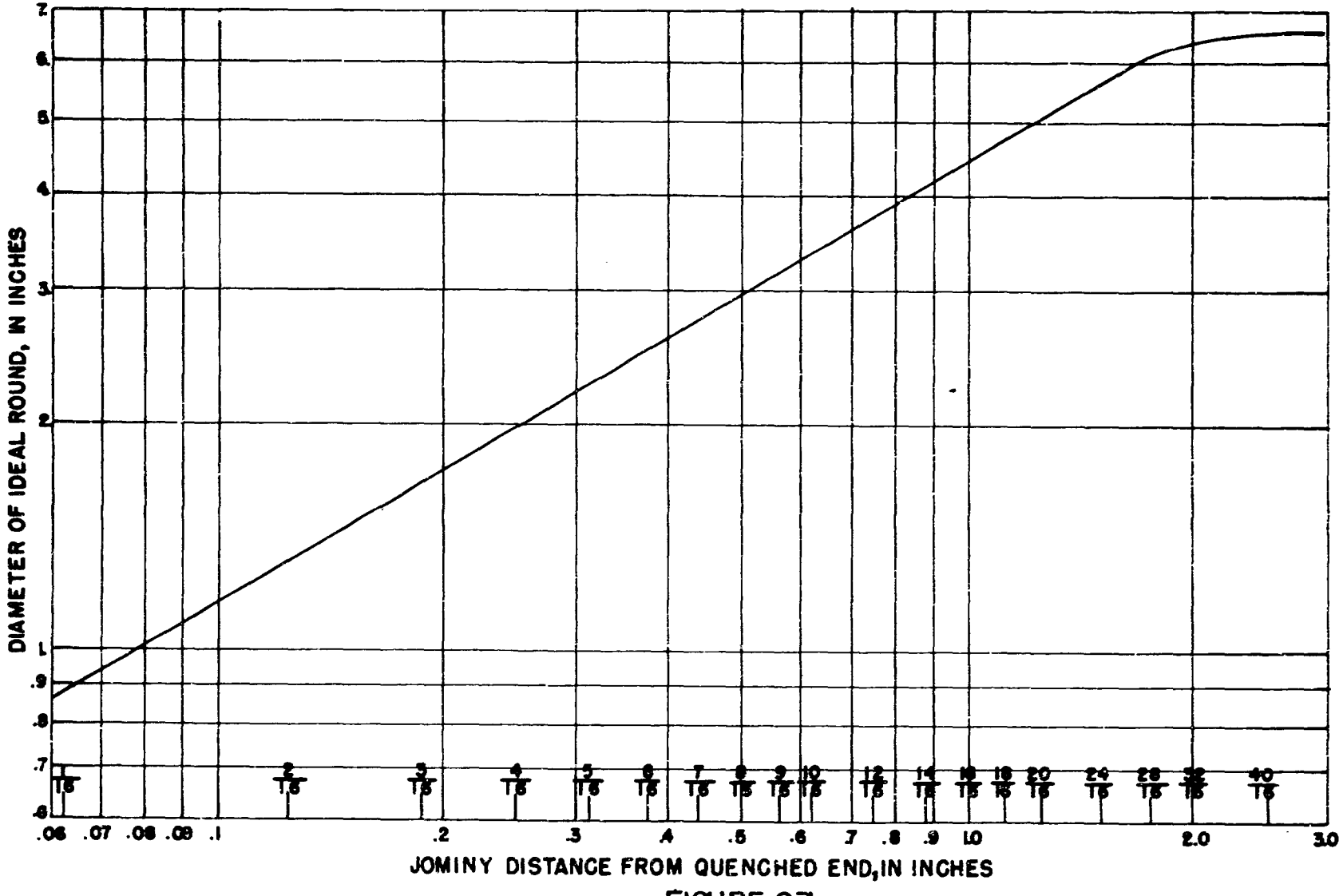

FIGURE 97

JOMINY DISTANCE AND IDEAL ROUND SIZE HAVING EQUIVALENT COOLING CONDITIONS
(DATA OF ASIMOW, CRAIG AND GROSSMANN BELOW 2 INCH JOMINY DISTANCE)

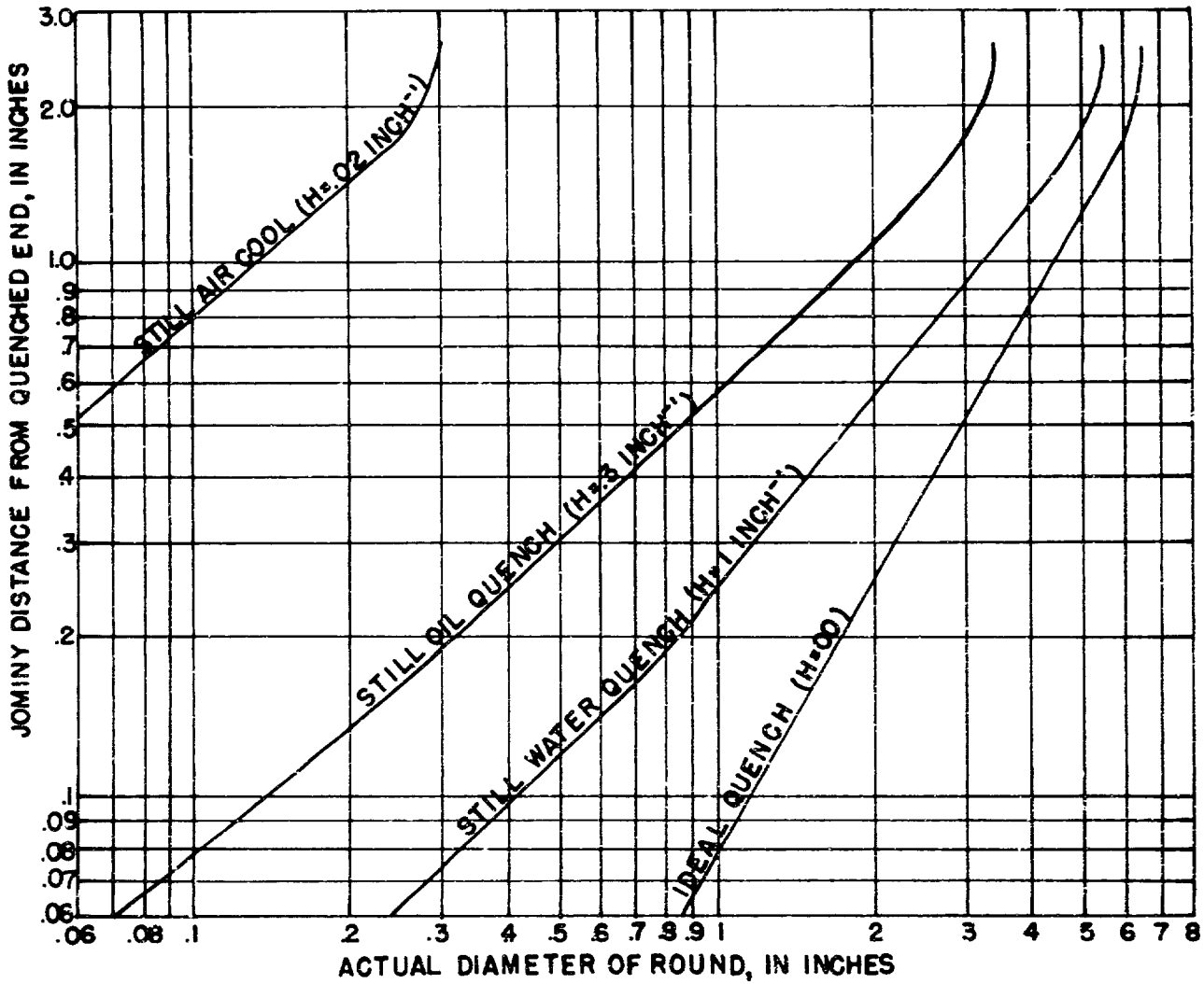

FIGURE 98

ROUND SIZE & JOMINY DISTANCE HAVING EQUIVALENT COOLING CONDITIONS

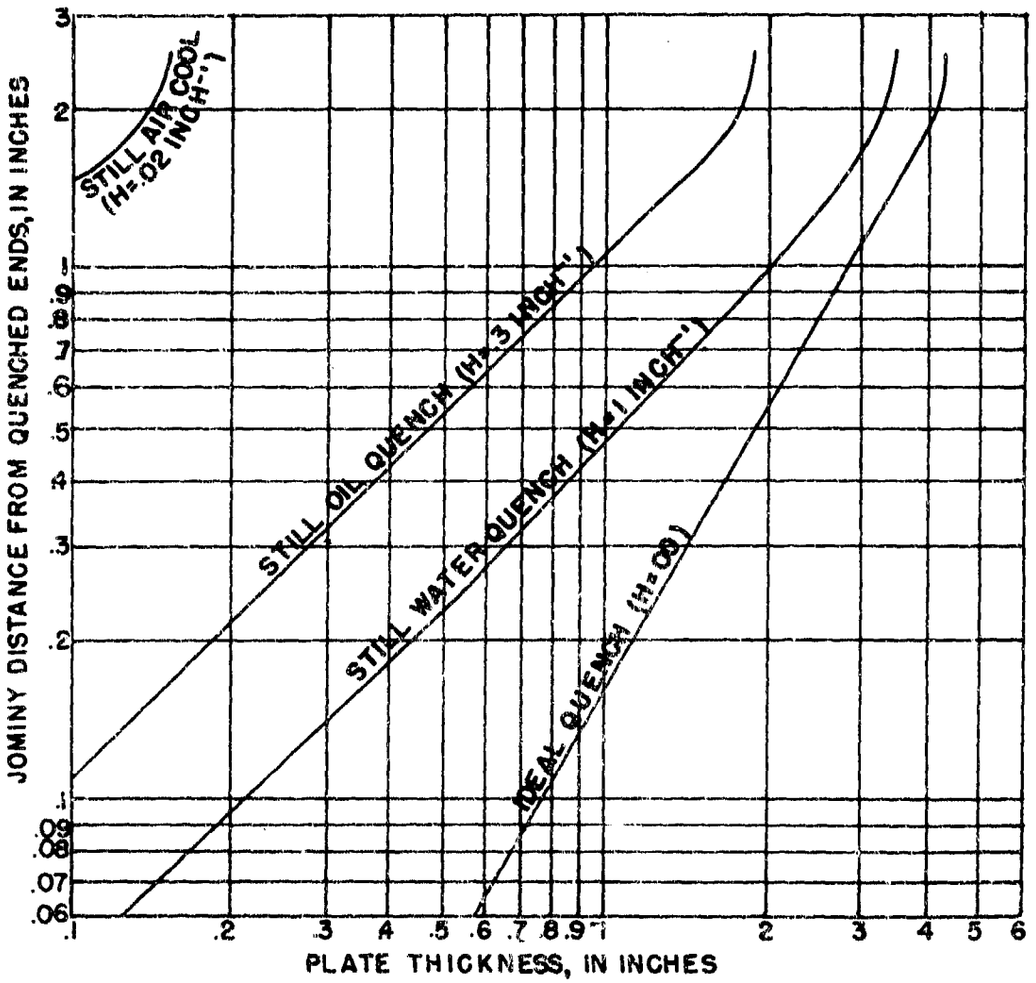

FIGURE 99

PLATE SIZE & JOMINY DISTANCE
HAVING EQUIVALENT COOLING CONDITIONS

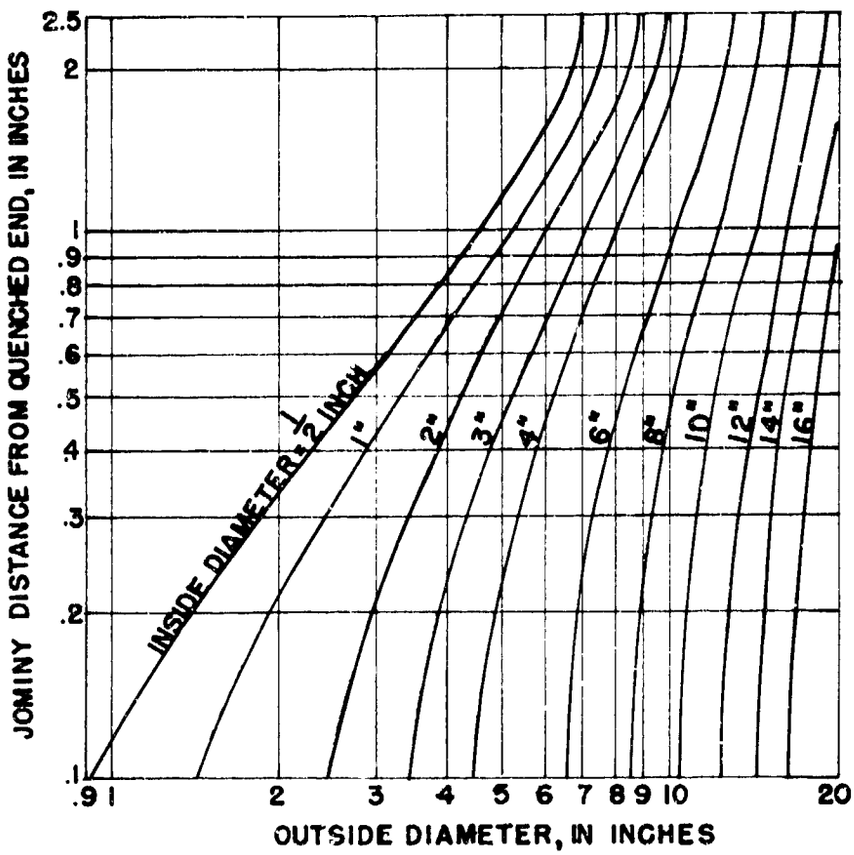

FIGURE 100

HOLLOW CYLINDER SIZE & JOMINY DISTANCE HAVING EQUIVALENT COOLING CONDITIONS.

CYLINDER WATER QUENCHED INSIDE & OUTSIDE ($H = 1\ \text{INCH}^{-1}$)

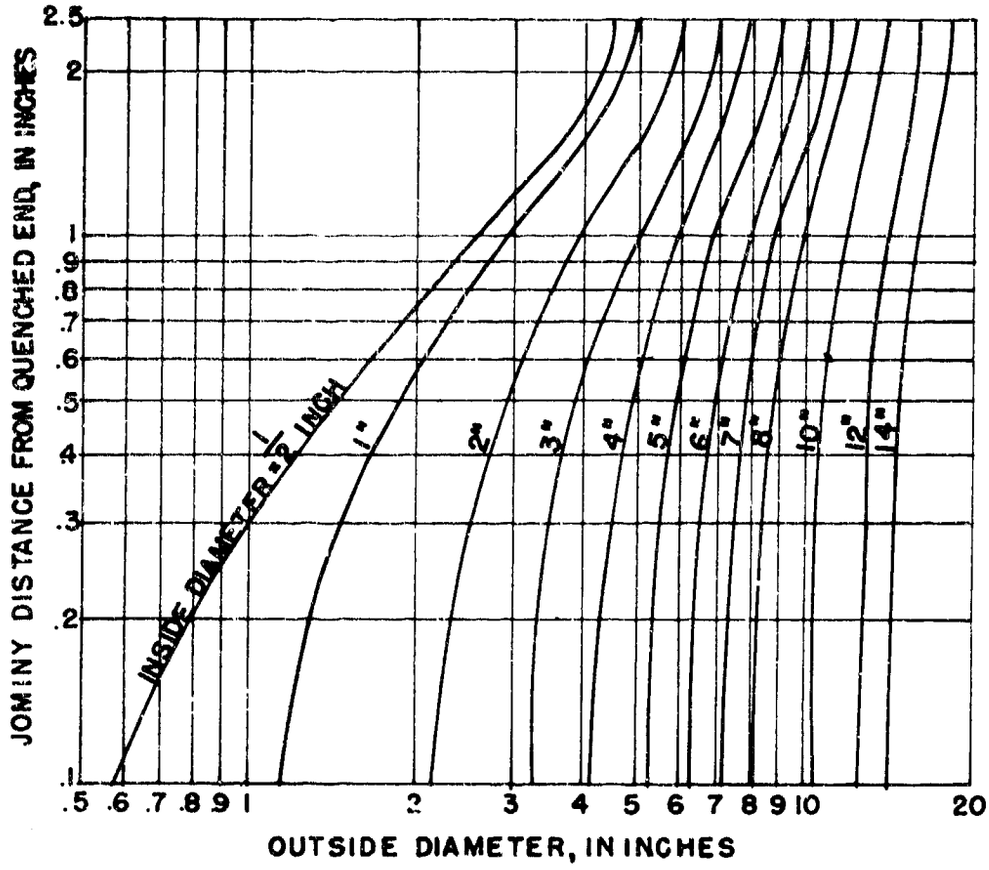

FIGURE 101

HOLLOW CYLINDER SIZE & JOMINY DISTANCE HAVING EQUIVALENT COOLING CONDITIONS.

CYLINDER OIL QUENCHED INSIDE & OUTSIDE ($H = .3$ INCH^{-1})

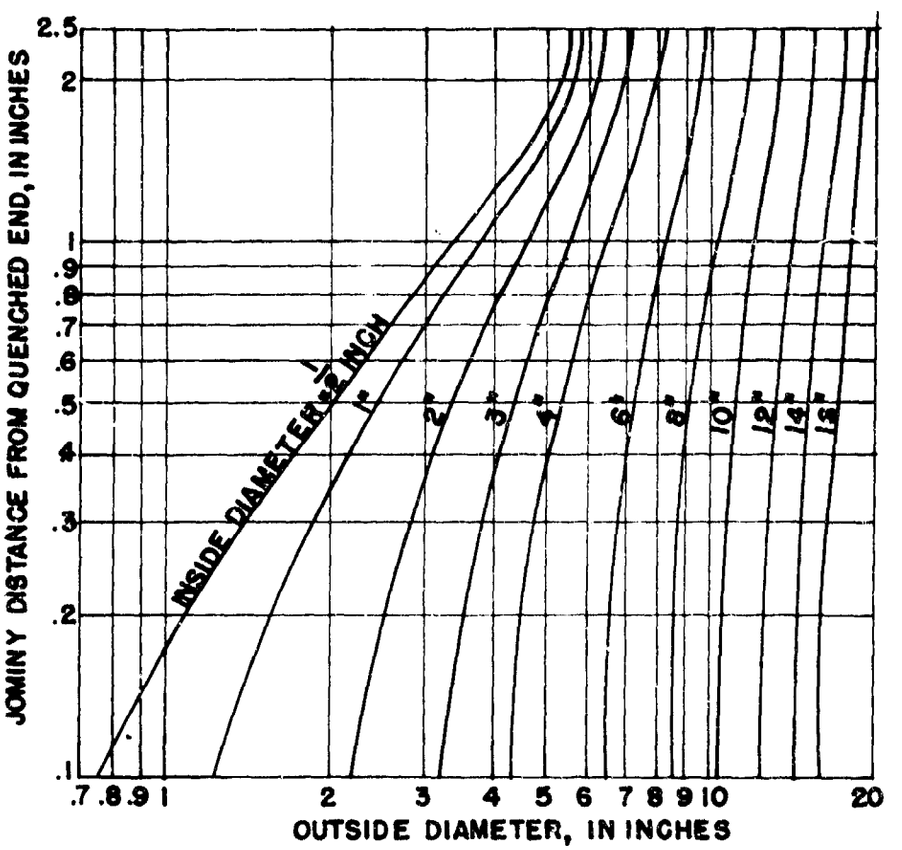

FIGURE 102

HOLLOW CYLINDER SIZE & JOMINY DISTANCE HAVING EQUIVALENT COOLING CONDITIONS

CYLINDER WATER QUENCHED OUTSIDE ONLY ($H = 1$ INCH^{-1})

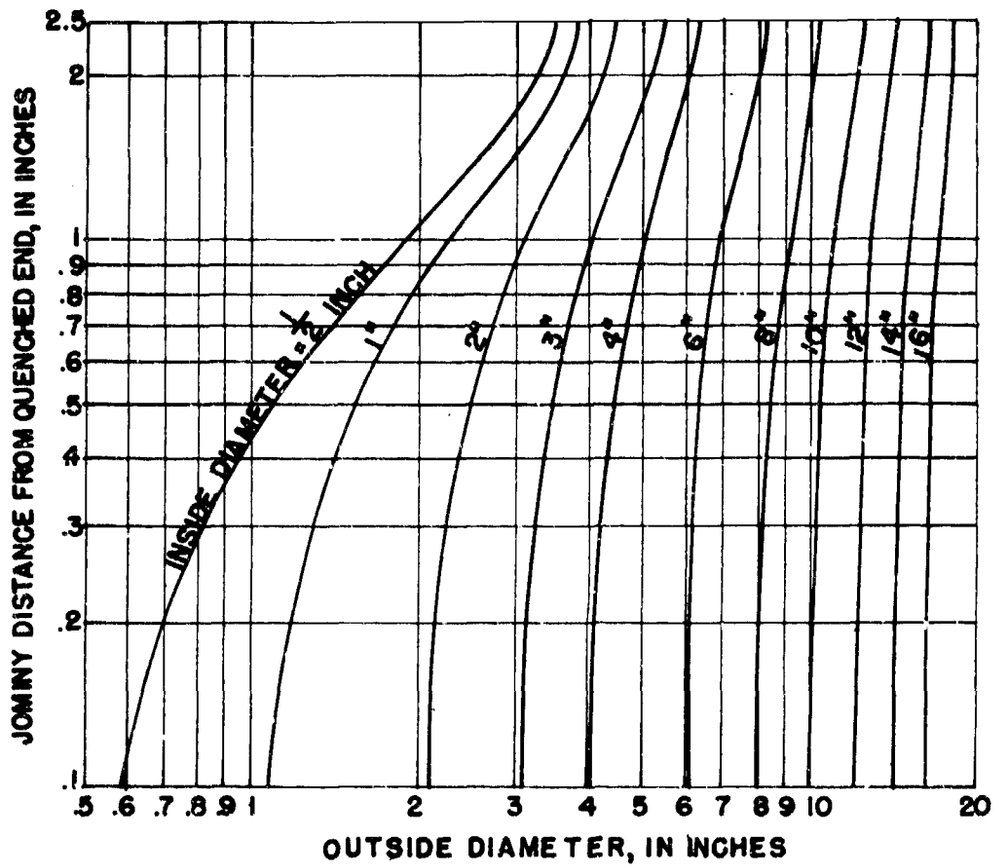

FIGURE 103

HOL OW CYLINDER SIZE & JOMINY DISTANCE
HAVING EQUIVALENT COOLING CONDITIONS

CYLINDER OIL QUENCHED OUTSIDE ONLY ($H = .3\ \text{INCH}^{-1}$)

Suggested Reading

N. B Pilling and T. D. Lynch:
"Cooling Properties of Technical Quenching Liquids", Trans. Am. Inst. Mining & Metal. Eng., 72 (1920) 665-688.

W. Scott:
"Quenching Media", Metals Handbook, Am. Soc. Metals, Cleveland, 1939, pp. 328-335.

VI. HARDENABILITY

If steel parts are to have a martensitic structure after quenching, they first must be austenitized and then cooled sufficiently rapidly to avoid the formation of pearlite, bainite, and proeutectoid products. If, on cooling, these transformations are avoided, the resulting structure can only consist of martensite (and retained austenite). The "hardenability" of a steel is a measure of the severity of cooling conditions necessary to avoid transformation in the pearlite and bainite ranges on continuous cooling(267). The less severe are the cooling conditions (the lower the rates of cooling) necessary to prevent the formation of pearlite and bainite, the greater is the hardenability. Since the pearlite and bainite reactions are independently controlled by the alloying elements, the duality of the hardenability concept must be recognized. The pearlitic hardenability is a measure of the severity of the cooling conditions necessary to avoid pearlite formation. The bainitic hardenability is a measure of the severity of the cooling conditions necessary to avoid bainite formation. The lower the maximum cooling rate at which pearlite forms upon continuous cooling, the greater is the pearlitic hardenability; and similarly for the bainitic hardenability. If, for a given steel, the pearlitic hardenability is less than the bainitic, the formation of pearlite will restrict the range of cooling conditions over which martensite will form. If the bainitic hardenability is less than the pearlitic, the formation of bainite will restrict the formation of martensite*.

Hardenability Measurements

To measure the hardenability it is necessary to quench a steel under a wide range of cooling conditions. The condition that is just sufficient to avoid both pearlite and bainite formation (or to form a specified percentage of bainite or pearlite) is then determined. If for slightly less severe conditions the product that forms is pearlite, the pearlitic hardenability has been determined, while if the product is bainite, the bainitic hardenability has been measured. If the bainitic hardenability limits the formation of martensite, the less severe conditions at which pearlite forms can also be measured.

Hardenability is most often measured by determining the maximum distance from the quenched end of a Jominy bar(259, 264, 268) at which the pearlite and bainite transformations do not occur, or occur

*It is possible that the proeutectoid ferrite and carbide reactions should be considered separately from the pearlite reaction, and that there are also ferrite and carbide hardenabilities. Little information is available on which to base judgement of such a procedure. In the remainder of this chapter, no distinction will be drawn between the pearlite and proeutectoid reactions, and references to pearlite formation will be understood to apply also to formation of proeutectoid products.

only to a specified extent. Hardenability can also be measured by determining the maximum size of round(65, 262) that contains no pearlite and bainite, or only a specified amount, when quenched in a medium of known severity. (Methods for determining the severity of quenching have been referred to in Chapter V.) The size of this round may be converted to an equivalent ideal round size by the chart of Figure 91. A number of other more or less specialized methods for measuring hardenability have also been proposed(269-277). Hardenability tests for low hardenability steels have been standardized by the Society of Automotive Engineers and the American Society for Testing Materials(264). The air-cooled end of a Jominy bar is equivalent only to the center of a 3-1/2 inch plate or a 5-1/2 inch round quenched in still water, and to a 2-inch plate or a 3-1/2 inch round in still oil (Figures 98 and 99). Where larger sections are of interest, the standard Jominy test is not applicable. No tests for high hardenability steels have as yet been generally adopted, but particular tests can be used to fit the part under study. The size of round that fully hardens can always be determined. However, high-hardenability steels are usually of interest in connection with thick sections and high HD values, for which the shapes of the cooling curves vary so greatly with the shape of the part and the position in the part that comparisons of cooling conditions are difficult.

The various methods of measuring hardenability may be related to each other through the criteria of equivalent cooling conditions discussed in the preceding chapter. Figure 97, for example, shows the relation between ideal round size and Jominy distance, on the basis of half-temperature time. It is well to remember that the hardenability is an intrinsic property of the steel and it is only the choice of the measure of hardenability that is arbitrary.

Generally it is desired to know the minimum rate of cooling (or some other measure of cooling conditions) that will completely prevent the formation of bainite and pearlite, for in many heat-treated parts it is desirable to produce completely martensitic structures. However, in practical tests of the hardenability of steel it has not been considered convenient to measure the maximum Jominy distance or the maximum round size for which the steel is completely martensitic. The presence of a small amount of bainite or pearlite is difficult to determine either by direct microstructural studies or by indirect mechanical property tests. It is known that the initiation of the pearlite transformation, at least, is not sensitive to changes in cooling rate. At about 50% pearlite and 50% martensite, small changes in cooling rate induce the greatest changes in structure(262). For this reason, "half-hardening" has been used as a criterion of hardening by many investigators. Frequently, macroetch, fracture, or hardness tests are used to delineate the half-hardened position. It has been found that in the case of martensite pearlite mixtures the macroetch(262, 269) and fracture tests(254, 269) in the as-quenched condition very nearly delineate the half-hardened boundary. The point of inflection of the hardness vs. Jominy distance curve (on a linear or semilogarithmic scale) and the point of inflection of the center hardness vs. diameter of round curve(262) also are nearly equivalent to the half-hardened structure, when the non-martensitic constituent is pearlite.

Other more or less arbitrary criteria of hardening, intended to represent half or full-hardening, are frequently used. In some cases, the cooling conditions that produce an arbitrary hardness decrease (such as 5 points Rockwell C) below the maximum hardness obtainable in the steel are considered to produce hardening. Hardness values, defined as a function of carbon content, are also used to determine the fully-hardened(264, 278) and half-hardened(262) conditions. Such definitions are of different significance(65, 67) when the non-martensitic constituent is bainite(279) than when it is pearlite(65).

Effect of Alloying Elements on the Pearlitic Hardenability

If a plain-carbon steel of moderate or small grain size is cooled at such a rate that it transforms to 50 per cent martensite and 50 per cent non-martensitic product, the non-martensitic product is pearlite, according to general belief. Therefore, if elements that retard equally the pearlite and bainite reactions (or that retard the bainite reaction more than the pearlite) are added to plain-carbon steels, the limiting hardenability of the resulting steels, on the basis of 50 per cent martensite, will be the pearlitic hardenability. If, on the other hand, alloying elements are added that have a greater retarding effect upon the pearlite transformation than on the bainite, a composition will be reached in which bainite will restrict the formation of martensite. Measurements of the effect of further additions of the alloying element will then apply to bainitic hardenability.

Several investigators(65-68) studied the effects upon hardenability of additions of various elements, singly, to essentially plain-carbon steels. Criteria intended to represent half-hardening were employed. For the additions of non-carbide-forming elements (which, as described in Chapter I, appear to retard the isothermal pearlite and bainite reactions equally), it appears that the measurements were of pearlitic hardenability. The measurements of the effects of small quantities of carbide-forming elements also presumably refer to pearlitic hardenability. The conclusion was reached(65) that the presence of a definite amount of each alloying element increased the (pearlitic) hardenability, expressed in terms of ideal round size for half-hardening, by a definite percentage. Since this was found to hold, approximately, regardless of the initial composition and hardenability, it was possible to establish multiplying factors for the effects of the different elements upon pearlitic hardenability. The pearlitic hardenability can, therefore, be calculated approximately by multiplying a base value by a factor for the amount of each alloying element present. This base value represents the hardenability (expressed in ideal round size) of a pure iron-carbon alloy.

The factors for manganese, nickel, molybdenum, chromium, silicon, phosphorus, and sulphur, selected(280) from several sources(65-68), and for copper(68), are indicated in Figures 104 to 106. Factors for boron (65, 280), zirconium(67), titanium(68), vanadium(65, 271, 67), aluminum (65, 66, 68), columbium(68), cobalt(68), beryllium(68), antimony(68),

arsenic(68), tellurium(68), tin(68), and germanium(68) may be found in the references indicated*. Since the multiplying factors for the alloying elements appear to vary linearly with the per cent of the alloying element (starting from 1.0 at zero per cent), they can be expressed in the form:

$$f = 1 + a\,x$$

where a is a constant for each element and x is the weight percentage of the element**. Table III gives the factors for the common elements in this form.

It was concluded(65) from other data(282) that the hardenability (ideal round size) varied as the square root of the carbon content, for hypoeutectoid steels. This relation was combined with the base value to give a base line for iron-carbon alloys, which may also be termed the multiplying factor for carbon (Figure 107 and Table III).

It was found(283,65) that a change in austenitic grain size of 1 ASTM number resulted in a fixed percentage change in (pearlitic) hardenability. The grain size factor can be combined with the base line for iron-carbon alloys, as shown in Figure 107. Table IV presents the same values in tabular form.

The hardenability base-line just mentioned refers to the size of ideal round that will half-harden at the center. Incomplete data(267,279) indicate that the hardenability required for half-hardening with respect to pearlite is about .75% of that required for full-hardening(267). Thus, the carbon factor (base-line) for full-hardening is approximately .75 times that for half-hardening, and is so given in Tables III and IV. The alloy factors derived for half-hardening appear to be valid for full-hardening also. The factors refer, of course, to elements dissolved in the austenite, not merely present in the steel, and are based upon an initial structure consisting of homogeneous austenite.

The form of the hardenability factors for carbon and the alloying elements indicate that the effectiveness of an element becomes less as the

*The effects of boron, zirconium, titanium, and vanadium upon hardenability are not well understood. The increase in hardenability produced by additions of these elements may be very great, but major discrepancies in the reported effects have been pointed out(267). It has been suggested (281) that these elements affect hardenability by reacting with nitrogen in the steel, and that their effect is, therefore, sensitive to the amount and form of nitrogen in the material before the addition of boron, zirconium, titanium, or vanadium.

**Because of this linearity it appears justifiable to use the factors for chromium and molybdenum for moderate percentages of these elements, even though they are based upon experimental data covering small percentages only.

per cent of the element increases. This decreasing effectiveness is illustrated in Figure 107, where the fractional increase in hardenability per additional percent of element added,

$$\frac{d(D_I)/D_I}{d(\%)}$$

is plotted against the weight per cent of the element. It is to be noted that carbon is the most effective of the common elements below .25 per cent, but above that, manganese is most effective.

This system for calculating hardenability from composition is empirical and has not been quantitatively related to the fundamental knowledge of austenite decomposition. The linearity of the factors for some of the alloying elements has been questioned(68) and there has been disagreement over some of the numerical values for the factors. Nevertheless, the system has been found to give the pearlitic hardenability for most hypoeutectoid steels within ± 10 per cent, and provides an approximate method for calculating hardenability that is often very useful.

Effect of Alloying Elements on the Bainitic Hardenability

Only recently has the duality of the hardenability concept become recognized(267), and there have been no investigations to determine experimentally the effects of alloying elements on the bainitic hardenability. A method of measuring the effects of the alloying elements on the bainitic hardenability has been suggested(267). Instead of adding the elements to plain-carbon steel it has been proposed that they be added to steels containing about .25% molybdenum. In steels containing this amount of molybdenum the formation of bainite restricts the formation of martensite on continuous cooling, and, therefore, the effects of the alloying elements on the bainitic hardenability can be measured. As yet no measurements of this type have been made. A tentative system of predicting the bainitic hardenability has been developed on the basis of a survey of published S-curves and Jominy test results, and may be of some use in cases where a rough estimate of the bainitic hardenability is required(267). In the derivation of this tentative system, it was assumed that the method of combining the effects of alloying elements for pearlitic hardenability applies to the bainitic hardenability also, and that changes in the rate of formation of pearlite or proeutectoid products do not in themselves alter the bainitic hardenability.

Nickel and manganese were, on the basis of isothermal transformation data, assumed to have the same effects on the bainitic hardenability as on the pearlitic. (Figure 104 and Table III.) Molybdenum was considered to have no effect on the bainitic hardenability. Chromium has some effect on bainitic hardenability but less than on the pearlitic. Because of the lack of data the assumption was made that chromium has one-half the effect on bainitic hardenability as that measured for pearlitic hardenability. Grain size was considered to have no effect. Changes in carbon content were considered to have the same effect on the bainitic as on the pearlitic

hardenability. The bainitic hardenability of iron-carbon alloys is, however, different from the pearlitic and thus a different value is necessary for the base-line for iron-carbon alloys. This base-line, together with the factors for the alloying elements, is given in Table III.

There is some evidence that the ideal round size for a trace of bainite is of the order of .5-.6 that for 50 per cent bainite(267,279). A value for the iron-carbon base-line based upon a ratio of .55 is given in Table III for use in calculating the size of ideal round that will fully harden with respect to bainite.

It is of interest to note that the ratio of the ideal round size for full hardening to that for half-hardening is considerably smaller when the non-martensitic product is bainite than when it is pearlite (about .55 as compared to approximately .75). As is indicated by Tables III and IV, it appears that for plain-carbon steels with medium austenitic grain size, the first trace of non-martensitic product obtained as the cooling rate is reduced below that necessary for full-hardening will be bainite. On the other hand, when the cooling rate is such that the structure is half non-martensitic, this half is primarily pearlite. The view that the first non-martensitic product is bainite in plain-carbon steels with medium or coarse grain is in contradiction to one recent report(282) and to common belief but there is some experimental evidence to support it(267, 284). The validity of the assumption that the published work on the effect of composition upon hardenability deals with pearlitic hardenability is not in question, since half-hardening rather than full-hardening was measured for in that work.

No measurements of the effects of composition on the hardenability of hypereutectoid steels appear to have been published, and there is some reason to believe that these effects may differ from those in hypoeutectoid steels. The values in Tables III and IV are, therefore, suggested at present for hypoeutectoid steels only.

It must be emphasized that the scheme suggested above for computing bainitic hardenability from composition is tentative and that the numerical values included in it are based upon scant data. The effect of such elements as titanium, boron, vanadium, and columbium cannot even be estimated at the present time. The system may serve, however, as the best approximation possible until further experimental data are available. The use of the factors in Table III is to some extent justified on the basis of the agreement between the measured bainitic hardenabilities and those calculated using these factors. The difference between the two for a number of steels has been found to be within ± 15 per cent(267).

Effect of Changes in Pearlitic Hardenability on the Bainitic Hardenability

In establishing the system for estimating the bainitic hardenability it was assumed that changes in pearlitic hardenability do not in themselves affect the bainitic hardenability. For example, since molybdenum does not appreciably alter the time for isothermal bainite formation, it was assumed that it does not affect the bainitic hardenability, even though molybdenum has a tremendous effect upon the pearlitic hardenability.

As indicated in the discussion (in Chapter I) of the relations among the stages of transformation, incubation in the pearlite temperature range affects the bainite reaction. If the first product to form in the pearlite range is ferrite, the greater the fractional time spent in this range, the shorter is the time required for the bainite transformation. If the first product to form is carbide, the greater the fractional time in the pearlite range, the longer is the time necessary for bainite formation. Thus, for low-carbon steels, increases in grain size, which have no effect upon the isothermal transformation to bainite, increase the bainitic hardenability by the indirect influence of less nucleation by ferrite formed in the pearlite range during continuous cooling(267). Increases in molybdenum content probably increase the bainitic hardenability similarly by decreasing the rate of ferrite formation. Such effects can not yet be taken into account quantitatively.

Hardenability of Commercial Steels

It is frequently necessary to use standardized commercial steels for parts designed to be fully hardened. Two series of such steels widely used in the United States are that of the Society of Automotive Engineers (S.A.E.) and the so-called National Emergency (N.E.) series.

The compositions of these steels are listed in the S.A.E. Handbook (285). The permissible ranges of the elements are rather wide, leading to a wide spread in hardenability between steels on the high side and on the low side of the ranges. Recently the trend toward purchasing steels on the basis of hardenability rather than composition has become more pronounced. Jominy hardenability specifications for some of the standard analyses have been developed, on the basis of the range of hardness vs. Jominy distance curves(286). These specifications are very wide and in many cases appear to be for austenitizing treatments that are insufficient for carbide solution. Furthermore, it is difficult to properly specify full-hardening in terms of hardness at a given point, the method prescribed in these specifications. The hardness is not only sensitive to variations in carbon content but is also affected, through retention of austenite, by alloy content and by cooling rate below the M_s temperature. The latter also influences the hardness through its effect upon the amount of tempering that occurs during the quench. The published curves concern, essentially, the lower of the pearlite and bainite hardenabilities only, and do not indicate which of these is being set.

Most of the S.A.E. and N.E. steels have low to moderate hardenability and will not provide full-hardening in any but the smallest sections. The pearlitic hardenabilities of some of the steels are adequate for rather heavy sections, but their bainitic hardenabilities are low. The pearlitic and bainitic hardenabilities calculated by the systems presented above, for the average compositions of the S.A.E. and N.E. steels, are given in Table V.

The wide spread in hardenability from high to low side analyses can be illustrated by an example. The table indicates that the pearlitic hardenability for full-hardening of the median analysis of S.A.E. 4340 is 10.3

inches. For the low side of the analysis it is 5.2 inches and for the high side, 11.4 inches. Most of the M.E. steels have relatively wider composition ranges and therefore relatively greater spreads in hardenability.

TABLE III

EFFECT OF CARBON AND ALLOYING ELEMENTS UPON HARDENABILITY (IDEAL ROUND)*

		Pearlitic Hardenability Factor	Bainitic Hardenability Factor
C	50 per cent martensite	$.338 \times \sqrt{\%\ C}$ in.**	$.494 \times \sqrt{\%\ C}$ in.
	Essentially all martensite	$.254 \times \sqrt{\%\ C}$ in.**	$.272 \times \sqrt{\%\ C}$ in.
Mn		$1 + 4.10 \times (\%\ Mn)$	$1 + 4.10 \times (\%\ Mn)$
P		$1 + 2.83 \times (\%\ P)$	$1 + 2.83 \times (\%\ P)$
S		$1 - .62 \times (\%\ S)$	$1 - .62 \times (\%\ S)$
Si		$1 + .64 \times (\%\ Si)$	$1 + .64 \times (\%\ Si)$
Cr		$1 + 2.33 \times (\%\ Cr)$	$1 + 1.16 \times (\%\ Cr)$
Ni		$1 + .52 \times (\%\ Ni)$	$1 + .52 \times (\%\ Ni)$
Mo		$1 + 3.14 \times (\%\ Mo)$	1
Cu		$1 + .27 \times (\%\ Cu)$	$1 + .27 \times (\%\ Cu)$

*After Hollomon and Jaffe(267).
**For grain size ASTM 7. For other grain sizes see Table IV.

TABLE IV

EFFECT OF AUSTENITIC GRAIN SIZE UPON PEARLITIC HARDENABILITY (IDEAL ROUND)*

Austenitic Grain Size (ASTM No.)	Carbon Factor (Inches) for Pearlitic Hardenability: 50 Per Cent Martensite	Carbon Factor (Inches) for Pearlitic Hardenability: Essentially All Martensite
1	.546	.410
2	.504	.378
3	.465	.350
4	.429	.323
5	.397	.298
6	.366	.275
7	.338	.254
8	.312	.234
9	.288	.217
10	.266	.200
	$\times \sqrt{\%\ C}$	$\times \sqrt{\%\ C}$

*After Hollomon and Jaffe(267).

TABLE V

CALCULATED HARDENABILITIES OF STANDARD

COMMERCIAL STEELS*

(IDEAL ROUND, IN INCHES)

Steel Number	Pearlitic Hardenability Full-Hard	Pearlitic Hardenability Half-Hard	Bainitic Hardenability Full-Hard	Bainitic Hardenability Half-Hard
SAE 1010**	.26	.34	.28	.50
1015	.32	.42	.34	.61
1016	.49	.65	.52	.95
1020	.36	.49	.39	.71
1022	.62	.82	.67	1.2
1024	1.1	1.4	1.1	2.1
1025	.41	.54	.44	.80
1030	.70	.93	.75	1.4
1035	.74	.99	.80	1.5
1036	1.2	1.6	1.3	2.4
1040	.80	1.1	.85	1.6
1045	.85	1.1	.91	1.7
1050	.90	1.2	.96	1.7
1052	1.4	1.9	1.5	2.8
1055	.93	1.2	1.0	1.8
1060	.97	1.3	1.0	1.9
1111***	.41	.55	.44	.80
1112	.39	.52	.42	.76
1113	.37	.49	.39	.71
1115	.55	.74	.60	1.1
1117	.76	1.0	.81	1.5
1118	.92	1.2	.99	1.8
1132	1.3	1.7	1.4	2.5
1137	1.4	1.8	1.5	2.7
1141	1.5	2.0	1.6	2.9
1145	1.0	1.3	1.1	2.0
1320	1.2	1.6	1.3	2.3
1330	1.4	1.9	1.5	2.8
1335	1.5	2.0	1.6	3.0
1340	1.6	2.2	1.8	3.2

*Compositions taken from 1944 SAE Handbook(285). Where a range is there specified for an element, the mean of the range was used. Phosphorus and sulfur, when no range was given, were taken as .030% each. Residual unspecified alloying elements were not considered. Grain size (for pearlitic hardenability) was taken as ASTM No. 7. Values based upon factors given in Table III.

**Silicon taken as .22% for the SAE 1000 series.

***Silicon taken as .05% for the SAE 1100 series.

TABLE V (CONT.)

Steel Number	Pearlitic Hardenability Full-Hard	Pearlitic Hardenability Half-Hard	Bainitic Hardenability Full-Hard	Bainitic Hardenability Half-Hard
NE 1345	1.7	2.3	1.8	3.2
1350	1.8	2.4	2.0	3.6
SAE 2317	1.1	1.5	1.2	2.2
2330	1.9	2.5	2.0	3.7
2340	2.4	3.2	2.6	4.7
2345	2.6	3.4	2.8	5.0
2515	1.4	1.8	1.4	2.8
3115	1.6	2.1	1.2	2.1
3120	2.3	3.0	1.7	3.1
3130	2.8	3.7	2.1	3.8
3135	3.0	4.0	2.3	4.1
3140	3.6	4.8	2.7	4.8
3141	4.1	5.4	2.9	5.3
3145	4.3	5.7	3.1	5.6
3150	4.5	6.0	3.3	5.9
3240	4.2	5.5	2.9	5.2
3310*	4.0	5.3	2.6	4.7
4023	1.2	1.6	.70	1.3
4027	1.3	1.7	.76	1.4
4032	1.4	1.8	.82	1.5
4037	1.6	2.1	.94	1.7
4042	1.7	2.2	1.0	1.8
4047	1.8	2.4	1.1	2.0
4119	2.3	3.0	1.0	1.8
4125	2.6	3.5	1.1	2.1
4130	2.8	3.7	1.2	2.2
4137	4.3	5.8	1.9	3.4
4140	4.8	6.4	2.1	3.7
4145	5.1	6.8	2.2	4.0
4150	5.3	7.1	2.3	4.2
4320	3.5	4.6	1.5	2.8
4340	7.7	10.3	3.1	5.7

*Phosphorus and sulfur taken as .015% each for SAE 3310.

TABLE V (CONT.)

Steel Number	Pearlitic Hardenability Full-Hard	Pearlitic Hardenability Half-Hard	Bainitic Hardenability Full-Hard	Bainitic Hardenability Half-Hard
SAE 4615	1.4	1.8	.83	1.5
4620	1.6	2.1	1.0	1.8
4640	2.7	3.6	1.6	3.0
4815	1.9	2.5	1.1	2.1
4820	2.5	3.3	1.5	2.7
5120	1.7	2.3	1.3	2.3
5140	2.5	3.3	1.8	3.2
5150	2.8	3.7	2.0	3.6
NE 8612	2.1	2.8	1.0	1.9
8615	2.4	3.2	1.1	2.1
8617	2.5	3.4	1.2	2.2
8620	2.8	3.7	1.3	2.4
8622	2.9	3.9	1.4	2.5
8625	3.1	4.0	1.5	2.7
8627	3.2	4.3	1.5	2.8
8630	3.4	4.5	1.6	3.0
8632	3.5	4.6	1.7	3.1
8635	3.9	5.2	1.9	3.4
8637	4.0	5.4	1.9	3.5
8640	4.2	5.5	2.0	3.6
8642	4.3	5.7	2.0	3.7
8645	4.4	5.9	2.1	3.9
8647	4.5	6.0	2.2	4.0
8650	4.7	6.2	2.2	4.1
8712	2.3	3.1	1.0	1.9
8715	2.6	3.5	1.1	2.0
8717	2.8	3.7	1.2	2.2
8720	3.0	4.0	1.3	2.4
8722	3.2	4.2	1.4	2.5
8725	3.4	4.5	1.5	2.7
8727	3.5	4.7	1.5	2.8
8730	3.7	4.9	1.6	3.0
8732	3.8	5.1	1.7	3.1
8735	4.3	5.7	1.9	3.4
8737	4.4	5.8	1.9	3.5
8740	4.6	6.0	2.0	3.6
8742	4.7	6.2	2.0	3.7
8745	4.8	6.4	2.1	3.9
8747	4.9	6.6	2.2	4.0
8750	5.1	6.8	2.2	4.1

TABLE V (CONT.)

Steel Number	Pearlitic Hardenability Full-Hard	Pearlitic Hardenability Half-Hard	Bainitic Hardenability Full-Hard	Bainitic Hardenability Half-Hard
NE 9255	2.0	2.7	2.1	3.9
9260	2.1	2.9	2.3	4.2
9261	3.0	4.0	2.7	5.0
9262	3.7	5.0	3.1	5.7
9415	1.9	2.6	1.2	2.1
9417	2.0	2.7	1.3	2.3
9420	2.2	3.0	1.4	2.5
9422	2.3	3.1	1.4	2.6
9425	2.5	3.3	1.5	2.7
9427	2.6	3.4	1.6	2.8
9430	3.0	4.0	1.8	3.2
9432	3.1	4.1	1.9	3.3
9435	3.2	4.3	1.9	3.5
9437	3.3	4.4	2.0	3.6
9440	3.4	4.6	2.1	3.8
9442	3.8	5.0	2.3	4.1
9445	3.9	5.2	2.4	4.3
9447	4.6	6.1	2.8	5.0
9450	4.7	6.3	2.8	5.2
9722	1.6	2.1	.90	1.6
9727	1.8	2.4	.99	1.8
9732	1.9	2.6	1.1	2.0
9737	2.1	2.8	1.2	2.1
9742	2.2	3.0	1.2	2.3
9745	2.3	3.1	1.3	2.3
9747	2.3	3.1	1.3	2.4
9750	2.4	3.2	1.4	2.5
9830	5.8	7.7	2.3	4.3
9832	6.0	8.0	2.4	4.4
9835	6.2	8.3	2.5	4.6
9837	6.4	8.5	2.6	4.7
9840	6.6	8.9	2.7	4.9
9842	6.8	9.1	2.8	5.0
9845	7.1	9.4	2.9	5.2
9847	7.2	9.6	2.9	5.3
9850	7.5	10.0	3.0	5.5
9912	2.3	3.1	1.0	1.9
9915	2.6	3.5	1.1	2.1
9917	2.8	3.7	1.2	2.2
9920	3.0	4.0	1.3	2.4
9922	3.2	4.2	1.4	2.5
9925	3.4	4.5	1.5	2.7

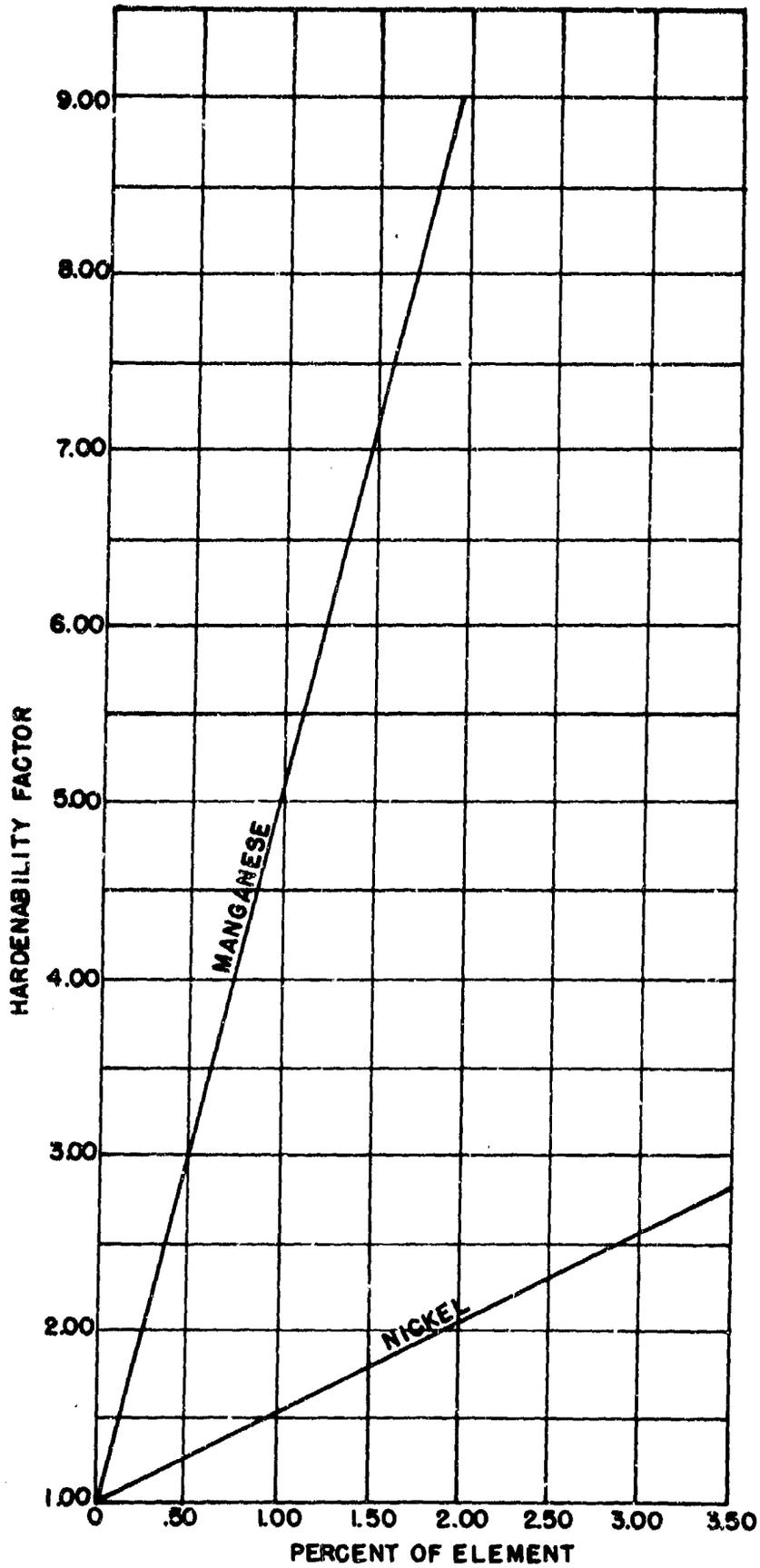

FIGURE 104
HARDENABILITY FACTORS FOR MANGANESE & NICKEL (AFTER GROSSMANN[60])

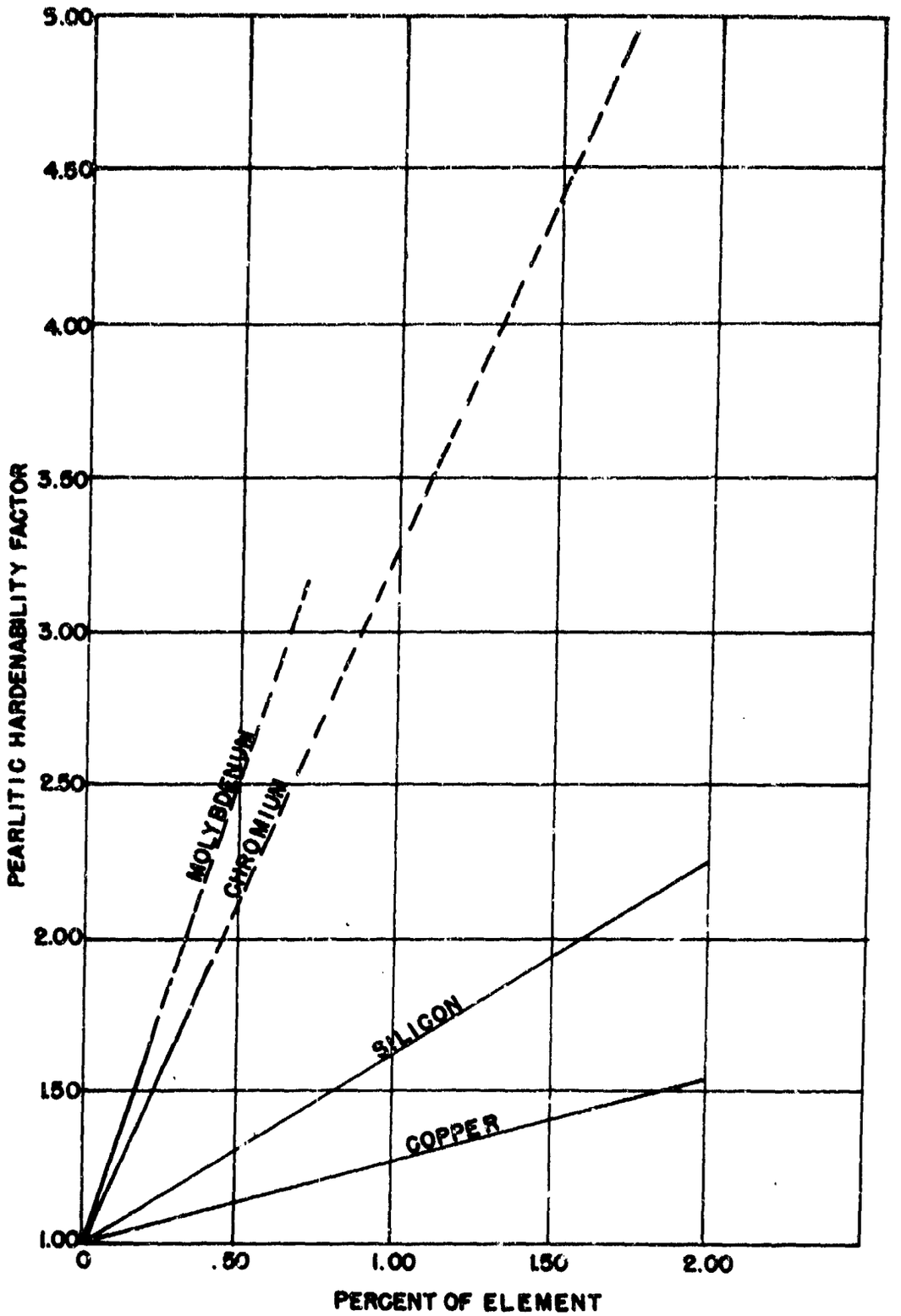

FIGURE 105

PEARLITIC HARDENABILITY FACTORS FOR MOLYBDENUM, CHROMIUM, & SILICON (AFTER GROSSMANN[250]) & FOR COPPER (AFTER KRAMER, HAFNER, & TOLEMAN[98]).

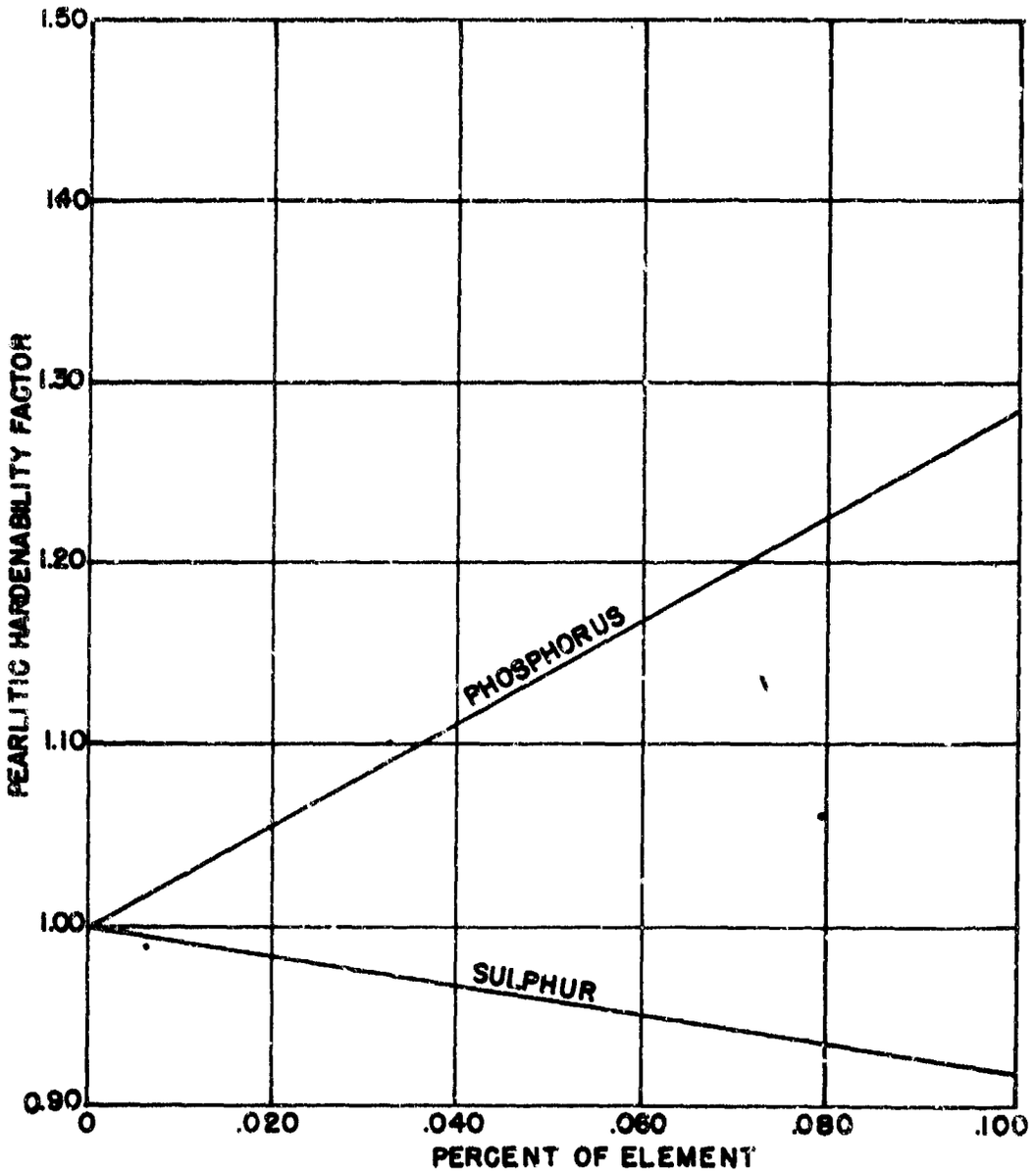

FIGURE 106

HARDENABILITY FACTORS FOR PHOSPHORUS & SULPHUR. (AFTER GROSSMANN[280]).

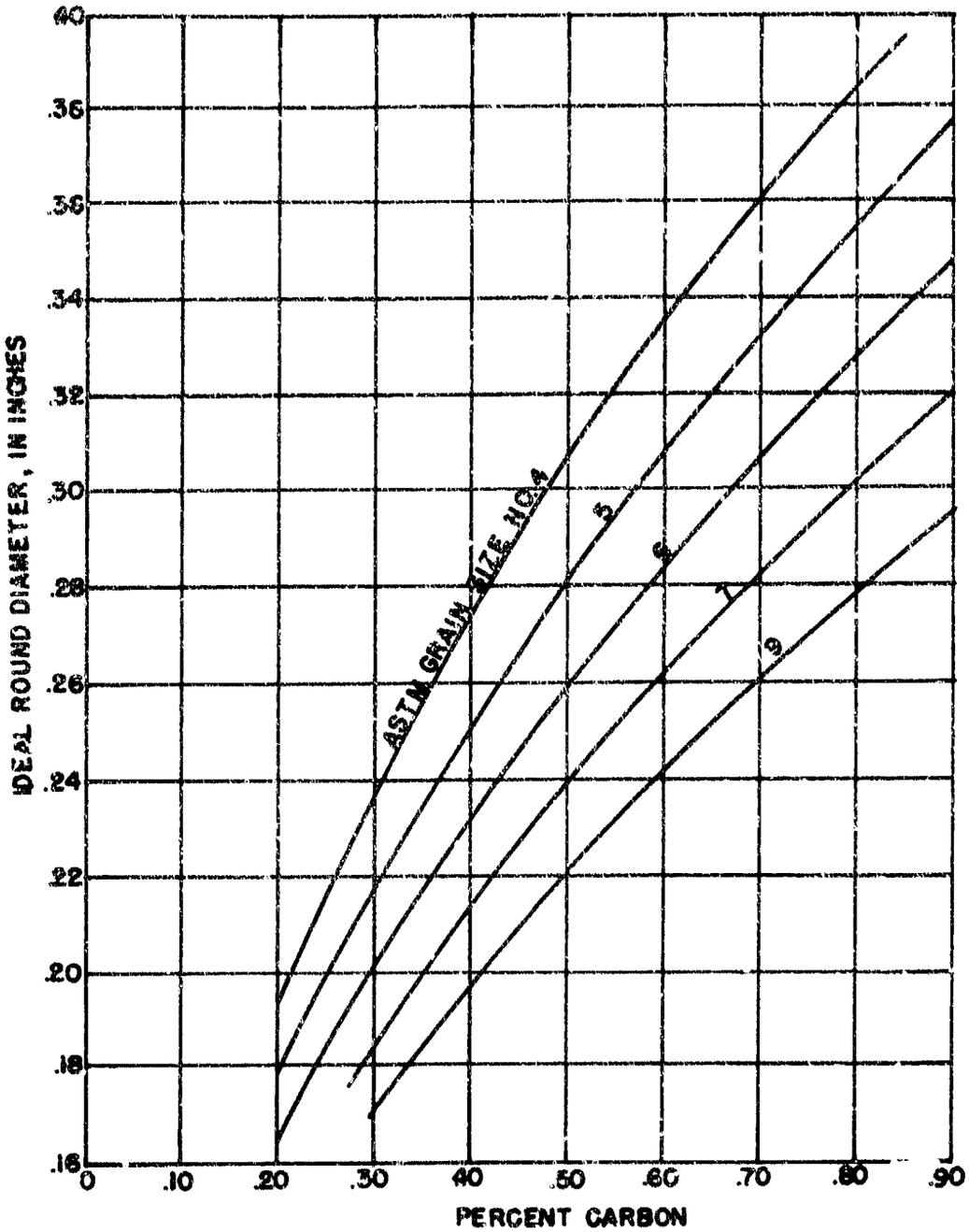

FIGURE 107

EFFECT OF CARBON CONTENT & AUSTENITIC GRAIN SIZE UPON PEARLITIC HARDENABILITY FOR HALF-HARDENING OF IRON-CARBON ALLOYS.
(AFTER GROSSMANN[20])

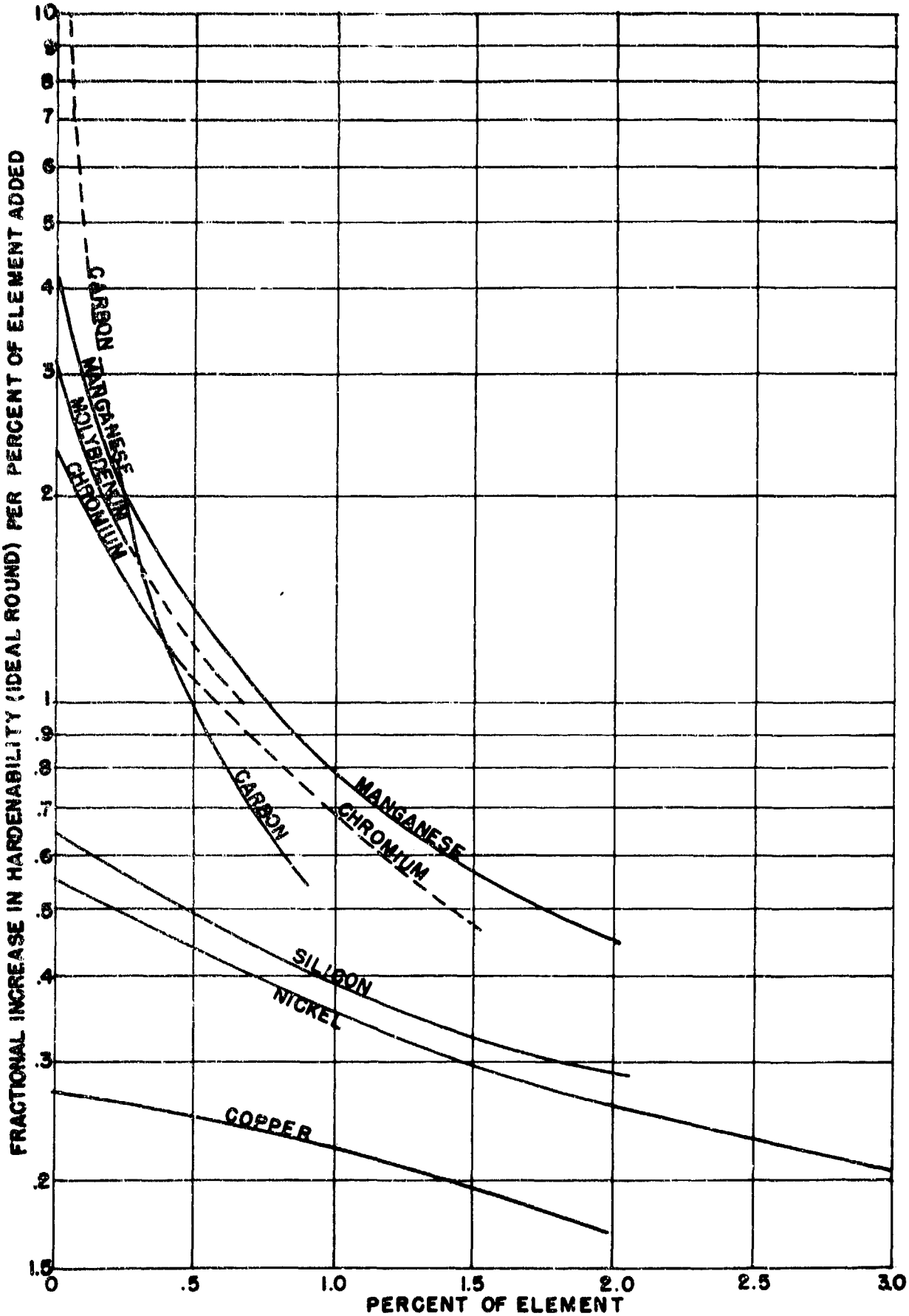

FIGURE 108

EFFECT OF INCREMENTAL ADDITIONS OF ALLOYING ELEMENTS UPON PEARLITIC HARDENABILITY

Suggested Reading

SAE Iron and Steel Division:
"Method of Determining Hardenability", SAE Handbook, 1943, pp. 315-321.
Also, Am. Soc. Testing Materials Std. A255-42T, 1942.

M. A. Grossmann:
"Hardenability Calculated from Chemical Composition", Trans. Am. Inst. Mining Metal. Eng., 150 (1942) 226-259.

J. H. Hollomon and L. D. Jaffe:
"The Hardenability Concept", Accepted for Publication, Am. Inst. Mining Metal. Eng., 1945.

VII. QUENCH-CRACKING

During the cooling of a steel part to produce martensite, temperature differences exist throughout the part. Because of these temperature differences, the contraction arising from the decreasing temperature and the expansion arising from the transformation of austenite to martensite do not occur uniformly over the part(287). Stresses are, therefore, set on a macroscopic scale. If the stresses exceed the flow stress of the material at the temperatures involved, flow will occur. Macroscopic flow during quenching results in a permanent change in shape, or "distortion". Such macroscopic flow also leaves stresses in the part that may cause further distortion upon subsequent reheating or machining. Heating lowers the flow stress and permits flow to take place under stresses insufficient to produce flow at room temperature. The removal of stressed material during machining necessitates a readjustment of the stresses in the remaining material; during this readjustment the flow stress is often exceeded locally, resulting in distortion. Stresses arising from quenching, if not relieved before the part is put into service, may have a significant effect upon its mechanical behavior.

If the macroscopic stresses set up by quenching are not relieved sufficiently by flow, they may exceed the fracture strength of the steel and cause "quench-cracking". Quench-cracking may occur during the quench or at some later time. Because of the localized shearing nature of the martensite transformation and perhaps also the heterogeneity of the macroscopic plastic flow, small regions of plastically deformed metal are set up that tend to relax in time. In the presence of a macroscopic stress (resulting either from an external load or from flow during quenching), relaxation of these regions may introduce stress concentrations that initiate fracture(103).

This gradual relaxation probably accounts for cracking that takes place some time after quenching. The minimum time necessary for appreciable cracking is of the order of days or hours at 32° F. (0° C.), hours or minutes at room temperature, and seconds at 212° F. (100° C.). At still higher temperatures, such as 350° F. (175° C.), no cracking occurs, presumably because the relaxation centers are removed. (Cracking may continue for years at room temperature and corresponding times at the other temperatures if the relaxation centers are not removed.)

Appreciable flow is more likely to occur the higher the macroscopic stresses and the lower the flow strength of the steel. Fracture is more likely to occur the higher the macroscopic stresses and the lower the fracture stress of the steel, relative to the applied stresses and to the flow stress. Fracture is also markedly affected by the microscopic stresses, but the factors affecting the formation of these stresses are not yet known. In almost all cases, a small amount of distortion is much less deleterious than cracking, for a cracked part is usually useless. It is, therefore, generally desirable to have the fracture stress high relative to the flow stress, as well as to have the macroscopic stresses as low as possible.

The macroscopic stresses responsible for quench-cracking in steel appear to arise primarily from the expansion accompanying the transformation of austenite to martensite, rather than from the thermal contraction accompanying the cooling of either austenite or martensite. Thermal contraction tends to set up compressive stresses near the surface of the part and tensile stresses in the interior. When the steel is hot, the flow stress is so low that little stress can be established. As the surface becomes cooler, it can withstand more stress. If the interior thereafter contracts (thermal contraction) more than the surface, the surface will be stressed in compression, and the interior in tension. If, however, after the surface becomes sufficiently strong, the interior on further cooling expands relative to the surface, the surface will be stressed in tension and the interior in compression. The interior can expand, on cooling, relative to the colder surface, only if the effect of the expansion accompanying the transformation from austenite to martensite is greater than the effect of normal thermal contraction. Fracture, as pointed out in Chapter III, arises only from tensile stresses. Since fracture starts most readily at the surface, where the likelihood of stress-concentrations is greatest, quench-cracking is most likely when the surface is left in tension by the quench; that is, when the effect of austenite-to-martensite expansion is greater than the effect of thermal contraction(268).

The occurrence of quench-cracking depends upon the separation between the flow-stress and fracture-stress curves (ductility) and upon the magnitude and direction of the macroscopic stress set up by quenching. These in turn, depend upon the size and shape of the part, the heat treatment (quenching practice) applied to the part, and the steel of which the part is made. These three factors will be discussed separately.

Effect of Part Shape

The temperature differences within the part are greater the greater the HD product (Chapter III). An estimate of the variation of surface and center axial stresses in a quenched cylinder with temperature differences and with HD is shown in Figure 109, for a hypothetical steel*. It will be noted that as the temperature differences (HD product) increase from zero, the tensional stress at the surface increases. As HD increases still further, the stress at the surface reaches a maximum, decreases, and finally changes to compression. The stress at the surface acts contrariwise. The change in stress occurs because at high HD values the center is so much hotter than the surface that the thermal contraction which must take place is greater than the expansion during the martensite transformation. As previously explained, this tends to put the surface in compression and the center in tension. Under such conditions, cracking is less likely than when the surface is tension. Over a certain range, then, increasing the thickness of the part and hence the HD product makes the stress distribution less likely to result in cracking. In most cases, however, the greater the thickness, the greater the likelihood of cracking.

*The tangential stress at the surface is approximately equal to the axial stress.

The shape of the part naturally affects the temperature differences developed within it. Sections of widely varying thickness tend to cool at widely varying rates, and wide temperature differences are likely to result.

Protruding corners also set up temperature differences. Since heat can flow from projecting corners to the quenching medium in several directions, the corners cool more rapidly than adjacent portions of the part. Moreover, circulation and turbulence of the quenching medium is often greatest near projecting corners. As a result of these factors an aqueous or oily medium is likely to commence rapid B stage cooling at projecting corners and thin sections considerably earlier than it does elsewhere, thus further increasing temperature differences along the surface.

Since heat flows during quenching to the base of a notch or re-entrant angle from several directions, these areas tend to be hotter than adjoining flat surfaces. Moreover, the circulation of the quenching medium is not as good over these areas as over adjacent ones (unless special sprays directed at the notches are used). As a result, if an aqueous or oil quench is used, the medium-steel interface at corners of notches and re-entrant angles is likely to remain in slow A stage cooling for a considerable time after the rest of the interface enters the B stage of more rapid cooling. This accentuates the temperature differences between the re-entrant angles and the remainder of the part surface.

Notches, besides affecting the macrostresses through changes in temperature distribution during quenching, act as stress-concentrators and so tend to raise the macrostresses in local regions. In this way notches make distortion and cracking more likely. Also, as discussed in Chapter IV, notches, by introducing transverse constraints and increasing the effective strain rate, raise the flow stress relative to the fracture stress. This raising of the flow stress facilitates fracture and makes cracking more likely. Ordinarily, the sharper, deeper, and longer the notch, the greater is its effect upon quench-cracking.

In general, then, quench-cracking and distortion are likely to occur if the part is thick, contains adjacent sections of widely varying thickness with sharp transitions, and has sharp corners either protruding, or particularly, re-entrant, so that they form notches. Excellent examples of such shapes have been published(289).

The shape effective during quenching may be modified without changing the shape of the rough part itself. For example, holes may be plugged with inserts made of steel or other metals. Threads may be covered with a plastic refractory. Thin sections adjacent to thicker sections may have thickening plates bolted over them. Such measures change the temperature distribution within the part and so change the macrostresses set up in the part. Since the added material is not ordinarily expected to withstand stress during quenching, it does not decrease the effects of a notch in concentrating the stresses and decreasing the separation between flow-stress and fracture-stress curves. Besides the additional work and expense involved in preparing and removing plugs and masks, their use sometimes has

the disadvantage of increasing the effective section thickness and retarding heat flow from steel to quenching medium. This is usually important only if the section to be masked or plugged is alread among the thickest on the part. In such cases, masking or plugging increases the hardenability required, and the necessary change in composition increases the susceptibility to cracking, and has other disadvantages, as will be discussed. Very often, however, masking and plugging are effective in eliminating quench-cracking and distortion.

Effect of Heat Treatment

The temperature differences within the part vary with the HD product, and so may be altered by changing the quenching practice. Reducing the severity of the quench (H) reduces the temperature differences and hence, in most cases, the quenching stresses. If the severity of quench were so low that the whole part was at a uniform temperature during the quench, the part would expand and contract as a whole, and no macroscopic stresses would be set up. Over a certain range, however, as was pointed out earlier, decreasing the temperature differences may result in a less favorable stress distribution, if the part is of uniform thickness. Here the temperature differences of interest are those involving regions that are transforming from austenite to martensite, and the pertinent quenching severity is that at surface temperatures corresponding to the martensite range within the part.

Provided the steel hardens, air cooling will generally result in lower macroscopic quenching stresses than will oil-quenching, and oil-quenching in lower stresses than water-quenching. However, such methods of decreasing the quenching severity in the martensite range also decrease the severity in the pearlite and bainite ranges, and thus increase the hardenability required. Consequently, the carbon or alloy content required to provide the hardenability is increased. Increases in carbon content decrease the resistance of the part to brittle failure (after heat-treatment), and, as will be discussed, in themselves increase the susceptibility to quench-cracking. Increases in alloy content generally increase the susceptibility to temper brittleness, as well as the cost of the steel. Increases in either carbon or alloy content generally lower the martensite temperature range, which, as will be shown, increases the tendency towards quench-cracking and retention of austenite.

It is clear that for hardening without cracking, a single medium would generally serve if it provided a very high quenching severity (at least as fast as cold brine) in the bainite range and a low severity (of the order of air) in the martensite range. No medium having such characteristics is known. Water and oil do approach the desired characteristics in that they usually provide appreciably greater severities in the bainite than in the martensite range. This may be one of the prime reasons for their being so widely used. In order to approach more closely the characteristics of the desired medium, recourse is often had to the use of two media.

The quenching severity in the martensite range can be reduced without greatly affecting the severity in the pearlite and bainite ranges

by "interrupting" the quench before the steel passes through the martensite range(290, 291). One type of interrupted quenching, known as "timed quenching", consists of quenching the part in a medium such as water or oil, and withdrawing it when its average temperature has fallen below the bainite nose but is still above the martensite range(290). The part is then cooled slowly, ordinarily in air, through the martensite range. In this method of quenching fairly close control of the time in the first medium is required. Timed quenching is not well adapted to parts having sections of widely varying thickness, since the thin sections will be quenched through the martensite range before the thick sections pass the bainite nose of the transformation diagram.

Another form of interrupted quenching consists of quenching into a bath held at a temperature between the bainite and martensite ranges, withdrawing the part after it has reached (approximately) the bath temperature, and cooling slowly through the martensite range(291). Molten salt is the medium commonly used for the initial quench. This quenching practice has been termed "martempering"(292). Martempering, because of limitations of the size of salt pots available, has not yet been applied to large pieces. Timed quenching, on the contrary, has probably been used more widely for large pieces than for small, since the longer times usually involved with large pieces are easier to control manually. Martempering, unlike timed quenching, is suitable for parts having sections of widely varying thickness. However, because of the nature of the media used, martempering provides low quenching severities in the pearlite and bainite ranges (Table II). The hardenability required for martempering is probably greater than for quenching in oil and considerably greater than for quenching in water. Timed quenching, on the other hand, hardly affects the cooling of the part through the pearlite and bainite ranges and so requires little, if any, more hardenability than does direct quenching.

It is possible to combine martempering with timed quenching. For example, a part having both thick and thin sections may be quenched in cold water until the thin sections are approximately at the M_s temperature, then transferred to a salt bath slightly above the M_s and hold there until all sections reach the bath temperature, and finally cooled in air through the martensite range. This treatment is more suitable than timed quenching for parts having varying section thicknesses, and may require somewhat less hardenability than martempering.

In interrupted quenching, the slower the rate of cooling through the martensite range, the smaller will be the macrostresses. The minimum rate of cooling is limited by the requirement that transformation to bainite be avoided. Also, the slower the cooling, the more austenite is likely to be retained (if the part is cooled to a fixed temperature). The use of a more severe second quench (such as oil), instead of air, reduces the tendency towards retention of austenite and formation of bainite and decreases the time involved, at the expense of somewhat increasing the stresses*.

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*In some cases of interrupted quenching (particularly timed quenching) the part has been put into a hot tempering furnace immediately after withdrawal from the quenching medium. This practice does not result in hardening. To obtain a martensitic structure it is necessary to cool the steel through the martensite range before tempering.

The temperature differences in the part during quenching are affected not only by the severity of quench but also by the temperature of the medium, being less the hotter the medium. The temperature differences within the part are, approximately, inversely proportional to the difference between part and medium temperatures. An increase in medium temperature does not of itself significantly affect the cooling rates in the pearlite and bainite ranges (unless the medium temperature approaches the bainite range). However, because of the effect of medium temperature upon their film coefficient, aqueous media generally cannot be advantageously heated above room temperature. Quenching oil, however, can be heated to temperatures approaching the flash point. Molten salt can be used at temperatures above 300° F. (150° C.) and low melting alloys (such as Woods' metal) at both lower and higher temperatures. Such media do not provide quenching severities in the pearlite and bainite ranges comparable to those provided by (cold) water, and their use in place of water increases the hardenability required, often to a considerable extent, with the consequence discussed above. If the temperature of the medium is within the martensite range, it is necessary, in order to obtain a fully martensitic structure, to remove the part from the medium and cool it further in a second colder medium. As the temperature of the first medium is increased, this practice approaches martempering.

Temperature differences within the part in the martensite range are affected by the initial temperature of the part. An appreciable effect occurs, however, only if the HD value is sufficiently high and if the initial temperature is sufficiently low that all portions of the part do not begin to cool exponentially before the surface enters the martensite range. The surface is ordinarily the last portion of the part to begin to cool exponentially, and Figures 89 and 90 give an indication of the fractional temperature at which this occurs. For example, they indicate that for HD values of 2 or greater the surface does not begin to cool exponentially until the fractional temperature is below .4. If, in this case, the quenching temperature can be made sufficiently low that the fractional temperature of the M_s is much above .4, any further decrease in quenching temperature will decrease the temperature differences within the part in the martensite range. For thick parts and severe media, temperature differences approaching the difference between initial and medium temperatures may occur within the part. In such cases, changes in the initial temperature result in almost equal changes in the temperature differences.

Since the minimum austenitizing temperature is set by the necessity for obtaining homogeneous austenite, it is usually desirable when lowering the quenching temperature to do so without affecting the austenitizing temperature. "Delayed quenching" consists of cooling the part in some mild medium, such as air, from the austenitizing temperature to some temperature above the bainite range, and then quenching in a more severe medium. With almost all steels, the quench may be delayed until the part reaches a temperature within the upper portion of the pearlite range. If the steel has sufficient pearlitic hardenability, the quench may be delayed until the part is below the pearlite range (though still above the bainite). Sufficient pearlitic hardenability can ordinarily be obtained by using sufficient molybdenum, without introducing the disadvantages that are likely

to accompany high bainitic hardenability: loss of toughness, lowering of the martensite temperature range, and greater tendency toward temper brittleness*.

Delayed quenching generally cannot be applied advantageously to parts having sections of widely different thickness because, unless the initial cooling severity is extremely low, the thin sections will cool into the bainite range before the thick sections are much below the austenitizing temperature. However, a procedure known as "stepped quenching" may be used. This consists of a preliminary cooling to a temperature between the pearlite and bainite ranges in a medium held at that temperature, followed, after all portions of the part are close to the medium temperature, by quenching in a second, cooler medium. The initial medium may be air, salt, or lead. With salt or lead, considerably less pearlitic hardenability is required than with air. Stepped quenching in salt or lead also requires less pearlitic hardenability than delayed quenching. Salt or lead baths are not available in as large sizes as air furnaces, however.

Stepped quenching may be combined with delayed quenching. The part may be cooled in a mild medium until the thinnest sections approach the upper limit of the bainite range, then transferred to a furnace at about this temperature and held until the thickest sections approach the furnace temperature, and finally quenched in a colder medium. If the first medium is air and the second lead or salt, the procedure results in small temperature differences within the part in the pearlite range. However, this hardly seems to be a significant advantage. If the initial medium is cold air and the second medium is air at a temperature above the bainite range, slightly less pearlitic hardenability is required than for a stepped quench in hot air. For large heavy pieces with sections of varying thickness this might prove advantageous.

The initial temperature for the cooling through the martensite range may also be reduced by interrupted quenching. Thus, in a timed quench, a part may be withdrawn from the initial severe medium, held long enough for thermal differences within it to equalize, and then replaced in the original medium for cooling through the martensite range. For such a quench, the quenching severity in the martensite range is just as great as for a direct quench in the same medium. Nevertheless, if the HD value is high, the thermal differences within the part may be much less than for a direct quench, because the initial temperature (for cooling through the martensite range) is, effectively, the temperature of equalization, rather

*If the pearlitic hardenability is originally higher than necessary, the cooling rate in the pearlite range may be decreased without a corresponding increase in pearlitic hardenability. This, however, is equivalent to holding for longer fractional times in the pearlite (or ferrite) range and for low-carbon steels may, as previously discussed, tend to decrease the bainitic hardenability of the steel. A corresponding slight increase in the inherent bainite hardenability of the steel may be needed to compensate for this loss, when delayed quenching is used.

than the austenitizing temperature. Equalization throughout a uniform one-inch plate takes about ten seconds, throughout a uniform six-inch plate about six minutes, the time varying as the square of the thickness. Since in the time necessary for equalization the average temperature can fall only slightly, temperature equalization of uniform sections is ordinarily carried out in room-temperature air. The temperature within a part having sections of widely varying thickness will equalize only at the medium temperature, except in media providing extremely low severities. Therefore, for non-uniform sections, equalization is best carried out in hot media. This again leads to the use of martempering, instead of (or in combination with) timed quenching, for non-uniform sections.

Interrupted quenching may be used in combination with either delayed or stopped quenching, or both.

It should be remembered that occasionally decreasing the temperature differences changes the stress distribution in such a way as to increase the likelihood of cracking.

It may also be noted that if the part is hardened at the surface but not at the center, the surface will tend to be stressed in compression and the center in tension, for two reasons: First, the transformation at the center to bainite or pearlite may cease while the surface is still weak, and second, the amount of expansion accompanying the formation of bainite or pearlite is less than that accompanying the formation of martensite. Both of these effects reduce the amount the center expands after the surface becomes strong enough to support appreciable stress, and so make the final surface stress tend towards compression rather than tension(288).

As was pointed out above, a quenched part is more likely to crack if, after quenching and prior to tempering, it is permitted to stand for a long time at room temperature, or for a shorter time at temperatures up to 250° F. (125° C.). Such holding apparently permits local regions of deformed material to relax and set up stress concentrations that may initiate fracture. If a part must be kept without tempering, it should be kept cold (at ice temperature or lower, to prevent relaxation) or heated for a short time above 350° F. (175° C.) (to remove the relaxation centers) (103).

Effect of Steel

Both the macrostresses and the separation between flow and fracture stress curves are affected by the nature of the steel itself. The separation between flow and fracture stresses is greater the lower the carbon content, and low-carbon steels are much less susceptible to quench-cracking than are high-carbon steels. Inclusions, porosity, flakes, seams and other defects lower the fracture strength and so increase the tendency towards quench-cracking. If the austenite contains much dissolved hydrogen, the hydrogen may precipitate during or following the transformation of austenite to martensite, and lower the fracture strength. This

may result in increased quench-cracking, just as it increases cracking following welding(293). A large austenitic grain size, too, appears to lower the fracture strength, and seems to make quench-cracking more likely(294). Finally, the separation between flow and fracture strengths is markedly affected by temperature. As the temperature is raised, the flow stress decreases more rapidly than the fracture stress, and fracture becomes more difficult. The higher the temperature of transformation from austenite to martensite, the more likely are the resultant stresses to be relieved by flow rather than by fracture. For this reason and for others to be discussed later, steels having their martensite range at a high temperature are much less likely to quench-crack than steels with a low martensite range.

Steel composition might affect the macrostresses set up during quenching by changing the temperature differences and by changing the change in volume per degree temperature difference. Since the severity of quench (H) is inversely proportional to the thermal conductivity of the steel (Chapter II), a decrease in thermal conductivity will increase the quenching severity, and hence the temperature differences. The thermal conductivities of austenite and martensite undoubtedly do decrease as the carbon and alloy contents increase, but the difference in conductivity between the various medium-alloy steels is probably relatively small.

The heat evolved during the martensite transformation retards the cooling and so decreases the temperature differences. The quantity of heat evolved is probably not directly affected to a significant extent by the composition differences between the various alloy steels. While thermodynamic considerations indicate that the quantity of heat liberated may vary as the temperature of the martensite transformation is changed, the direction of the effect is doubtful and the effect itself probably small. Changes in the steel, then, are not likely to have an important influence upon the magnitude of the temperature differences during quenching.

The differences in volume are affected not only by the temperature differences but also by the change in volume per degree change in temperature. The coefficients of thermal contraction for austenite and martensite are only slightly affected by changes in composition within the usual ranges. The change in volume accompanying the transformation of austenite to martensite is, however, markedly, though indirectly, affected by composition. It was pointed out in Chapter I that the expansion accompanying the martensite transformation is practically independent of composition, for a fixed temperature of transformation. The amount of expansion does increase considerably as the temperature of transformation is decreased (Figure 16). Changes in composition that lower the martensite temperature range thus increase the differences in specific volumes within the part and so increase the likelihood of cracking. This, it must be remembered, is in addition to the effect of low martensite range in increasing cracking by raising the flow strength relative to the fracture strength.

Two factors fixed by the composition, therefore, have a significant effect upon the susceptibility to quench-cracking: the carbon content and the martensite temperature range. The problem of finding the composition that has the minimum tendency toward quench-cracking thus resolves itself into one of obtaining the composition that provides the desired hardenability with the maximum martensite temperature range, and with the minimum carbon content necessary to obtain the desired strength with the heat treatment procedure employed(118). The maximum martensite temperature at the same time provides the minimum amount of austenite retained after the quench, and so minimizes difficulties arising from retention of austenite*. An indication of the likelihood of quench-cracking and austenite retention in fully-hardened standard steels may be obtained from the martensite-start (M_s) temperatures tabulated in Table VI.

The composition having the maximum M_s temperature for a given hardenability may be obtained from information previously given concerning the effect of composition upon the M_s temperature and upon hardenability. (It might be more pertinent to quench-cracking to consider the entire martensite range rather than only the M_s temperature, but reliable data for the effect of composition upon the entire range are not available.) Values for the effects of various elements upon the M_s temperature are given in Table I. As indicated in Chapter I, the effects of alloying elements on the martensite-start temperature appear to be additive. To a first approximation, the pearlitic hardenability of hypoeutectoid steels may be calculated from composition on the basis of the multiplicative system for calculating hardenability, using the data given in Table III. As yet only a tentative system for calculating bainitic hardenability (Chapter VI) is available; assumed values for use in this system are also given in Table III.

To compare the relative effects of the alloying elements upon hardenability and on the martensite transformation temperature, it is necessary to determine the effect of each element on the fractional increase in hardenability per degree lowering of the martensite transformation. In Figure 110 is plotted the fractional increase in bainitic hardenability (ideal round) per degree centigrade lowering of the martensite-start temperature,

$$\frac{-d(D_I)/D_I}{d(M_s)},$$

for various amounts of the different alloying elements. Figure 111 is the corresponding plot for pearlitic hardenability. These figures are based upon the data presented in Tables I and III**. When more complete and accurate information becomes available, the new data may be used but the genera

*It may be noted that while decreasing the severity of quench decreases the tendency toward quench-cracking (if the composition is unchanged and the steel transforms completely to martensite), the resulting decrease in speed of cooling increases the amount of austenite retained.

**The special addition elements, such as vanadium, titanium, zirconium, and boron have not been considered for the reasons discussed in Chapter VI.

principles should remain unchanged. Since little is known concerning the effect of alloying elements on the hardenability of hypereutectoid steels, Figures 110 and 111 are intended to apply to hypoeutectoid steels only.

The higher is a point on a curve for an element in the figures, the greater is the increase in hardenability per degree lowering of the M_s temperature for further additions of that element. It will be noted that moderate additions of molybdenum increase the pearlitic hardenability with less lowering of the M_s temperature than do additions of manganese, chromium, nickel, copper, or carbon. Molybdenum increases the pearlitic hardenability greatly, and .30 per cent is sufficient to insure that the pearlitic hardenability is greater than the bainitic. Thus, it is only necessary to select the composition giving the desired bainitic hardenability with the maximum M_s temperature, and then add enough molybdenum to increase the pearlitic hardenability as far as may be needed. The choice of the combination of alloying elements other than molybdenum can, therefore, be based upon bainitic hardenability versus M_s. (Molybdenum itself, as has been pointed out, has very little direct effect upon bainitic hardenability.)

Figure 110 indicates that the minimum increase in bainitic hardenability for a given lowering of the martensite-start temperature is obtained with the addition of carbon (after the first .05 per cent). Thus, not only does a low carbon content provide maximum toughness of the quenched steel both before and after tempering, but also maximum martensite transformation temperature.

Since the addition of silicon does not affect the martensite transformation temperature (Table I), yet increases the bainitic hardenability (Table III), the higher this element the less the tendency toward quench-cracking, for constant hardenability. However, in cast steels, at least, it has been reported that silicon contents much in excess of .50% lower the impact properties(80). Possibly this decrease is due to the formation of certain silica base-inclusions. At least at present, the per cent of silicon that can be used without hesitation is limited to about .50 per cent.

Figure 110 indicates that of the three elements manganese, nickel, and chromium, manganese should, for minimum quench-cracking, be added first until about .20% is present. Then, since the next addition of manganese would lower the martensite-start temperature more than would an initial addition of chromium (for a given increase in hardenability), chromium should be added. But if chromium were added alone, the martensite-start temperature would again be unnecessarily lowered, for the chromium curve would drop to a lower value than that for .20% manganese. Therefore, after a steel contains about .20% manganese, chromium and manganese should be added together, and in such quantities that the points corresponding to their concentrations be on the same horizontal line in Figure 110. These concentrations are indicated in Figure 112. The same situation arises with regard to nickel; after the steel contains about .60% manganese and .80% chromium, nickel, chromium and manganese should be added together, and in such quantities that the points corresponding to their concentrations be on the same horizontal line in Figure 110. These concentrations are also indicated in Figure 112. Thus, according to the available information as to the effect of the alloying elements on the bainitic hardenability

and on the martensite-start temperature, manganese, nickel, and chromium should be added simultaneously and in concentrations bearing a definite relationship to each other. Instead, then, of considering individual hardenability factors for each element, it is possible to determine a combined factor for the three elements in the relative amounts given by Figure 112. This factor, plotted in Figure 113 as a function of the manganese content, is the product of the individual factors for the three elements, manganese, nickel, and chromium.

Figure 110 indicates that when sufficient hardenability is required, copper should be added simultaneously with manganese, chromium, and nickel. However, because copper has limited solubility in ferrite and causes age-hardening, it is not ordinarily considered as an alloying element for obtaining hardenability.

It is to be remembered that since the numerical data concerning bainite hardenability are so few, the numerical values given in Figures 110, 112 and 113 are necessarily tentative. It is the principles shown by these figures, rather than the numerical values, that should be kept in mind.

TABLE VI

CALCULATED MARTENSITE-START TEMPERATURES OF STANDARD COMMERCIAL STEELS*

Steel Number	M_s °C	M_s °F	Steel Number	M_s °C	M_s °F
SAE 1010	500	930	SAE 2317	410	770
1015	480	890	2330	355	680
1016	465	870	2340	320	600
1020	465	870	2345	300	580
1022	445	840			
1024	410	770	2515	390	740
1025	445	840			
1030	410	770	3115	445	830
1035	395	750	3120	420	780
1036	375	700	3130	385	790
1040	380	720	3135	365	690
1045	360	680	3140	345	650
1050	340	640	3141	340	650
1052	315	600	3145	325	615
1055	325	620	3150	305	580
1060	310	590			
			3240	340	640
1111	485	900			
1112	485	900	3310	405	760
1113	485	900			
1115	465	860	4023	440	820
1117	445	830	4027	420	790
1118	430	810	4032	405	760
1132	385	720	4037	380	720
1137	365	690	4042	365	690
1141	345	660	4047	345	660
1145	360	680			
			4119	440	820
1320	410	770	4125	415	780
1330	375	710	4130	405	760
1335	355	680	4137	370	690
1340	340	640	4140	355	670
			4145	335	640
NE 1345	320	610	4150	325	610
1350	305	580			

*Compositions taken as for Table V. Temperatures calculated from Centigrade values given in Table I.

TABLE VI (CONT.)

Steel Number	M_s °C	M_s °F	Steel Number	M_s °C	M_s °F
SAE 4320	415	780	NE 9255	325	620
4340	330	630	9260	305	580
			9261	305	580
4615	440	830	9262	300	570
4620	425	800			
4640	350	660	9415	440	830
			9417	435	820
4815	415	780	9420	425	800
4820	395	740	9422	420	780
			9425	405	760
5120	435	820	9427	400	750
5140	360	680	9430	385	730
5150	325	610	9432	380	710
			9435	370	690
NE 8612	455	850	9437	360	680
8615	445	830	9440	350	660
8617	435	820	9442	340	640
8620	425	800	9445	330	620
8622	420	790	9447	315	600
8625	410	770	9450	305	580
8627	400	760			
8630	390	740	9722	435	810
8632	385	720	9727	415	780
8635	370	690	9732	400	750
8637	365	690	9737	380	720
8640	355	670	9742	365	680
8642	345	660	9745	350	670
8645	335	640	9747	345	650
8647	330	620	9750	375	640
8650	320	600			
			9830	375	710
8712	455	850	9832	370	700
8715	445	830	9835	360	680
8717	435	820	9837	350	670
8720	425	800	9840	340	650
8722	420	790	9842	335	640
8725	410	770	9845	325	620
8727	400	750	9847	317	600
8730	390	740	9850	305	580
8732	385	720			
8735	370	690	9912	450	840
8737	365	680	9915	440	820
8740	350	670	9917	435	810
8742	345	660	9920	425	790
8745	335	640	9922	415	780
8747	330	620	9925	405	760
8750	315	600			

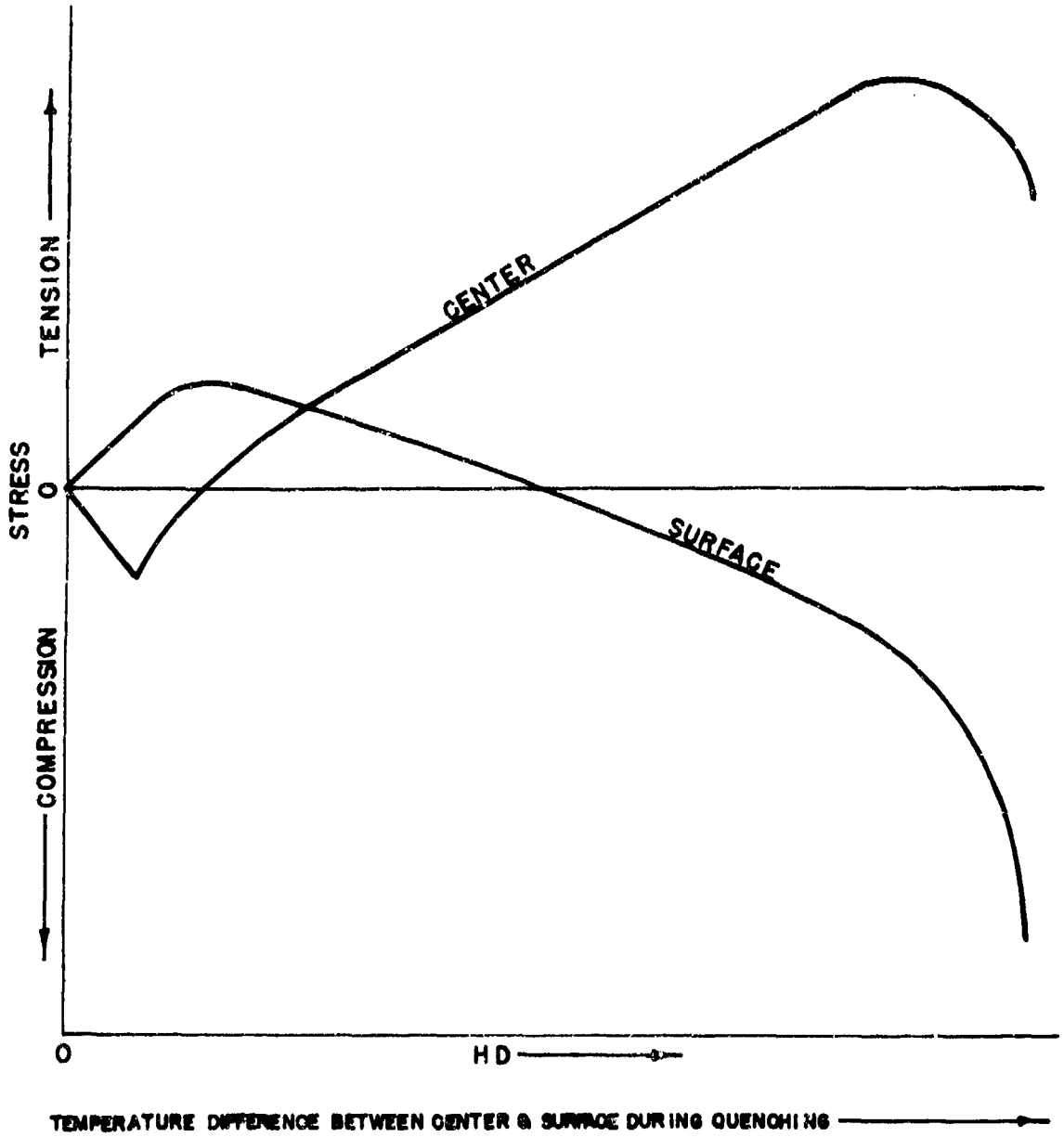

FIGURE 109

COMPUTED AXIAL STRESS WITHIN A CYLINDER OF A HYPOTHETICAL STEEL QUENCHED TO MARTENSITE. (FROM SCOTT[206]).

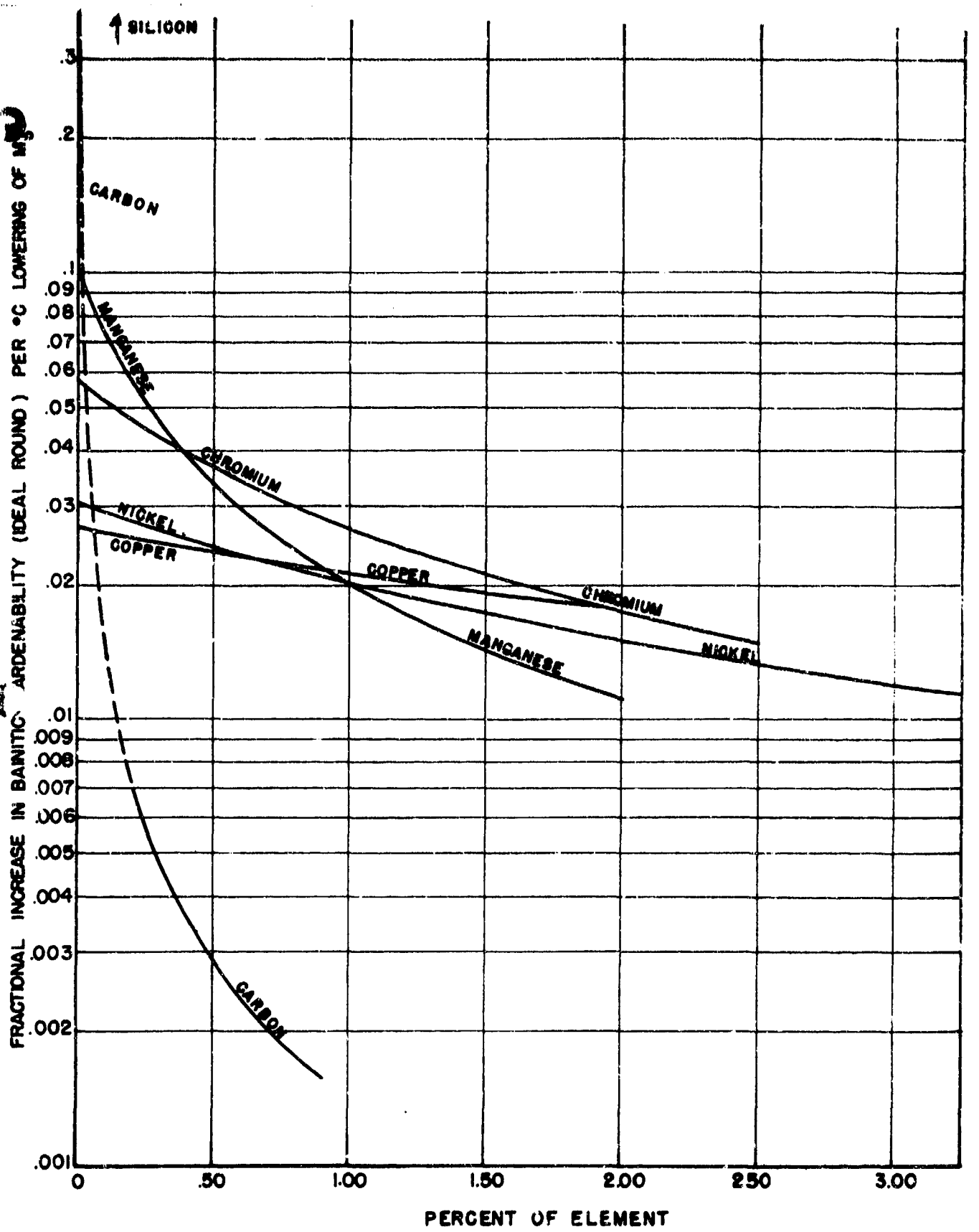

FIGURE 110

RATIO OF EFFECT OF ALLOYING ELEMENTS

UPON BAINITIC HARDENABILITY TO THEIR EFFECT UPON THE M_s TEMPERATURE

(AFTER JAFFE & HOLLOMON[118])

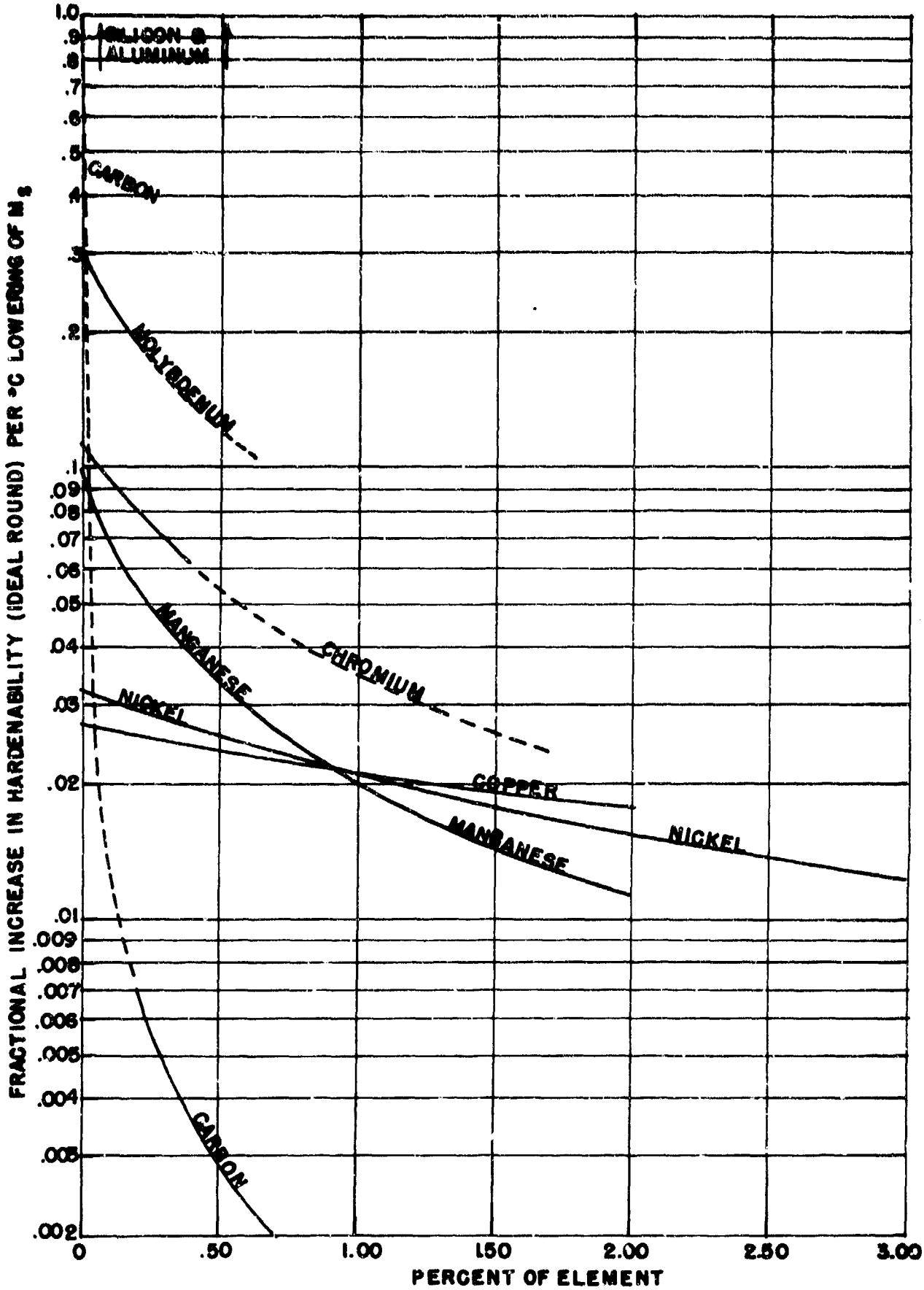

FIGURE 111

RATIO OF EFFECT OF ALLOYING ELEMENTS UPON PEARLITIC HARDENABILITY TO THEIR EFFECT UPON THE M_s TEMPERATURE (AFTER JAFFE & HOLLOMON[118])

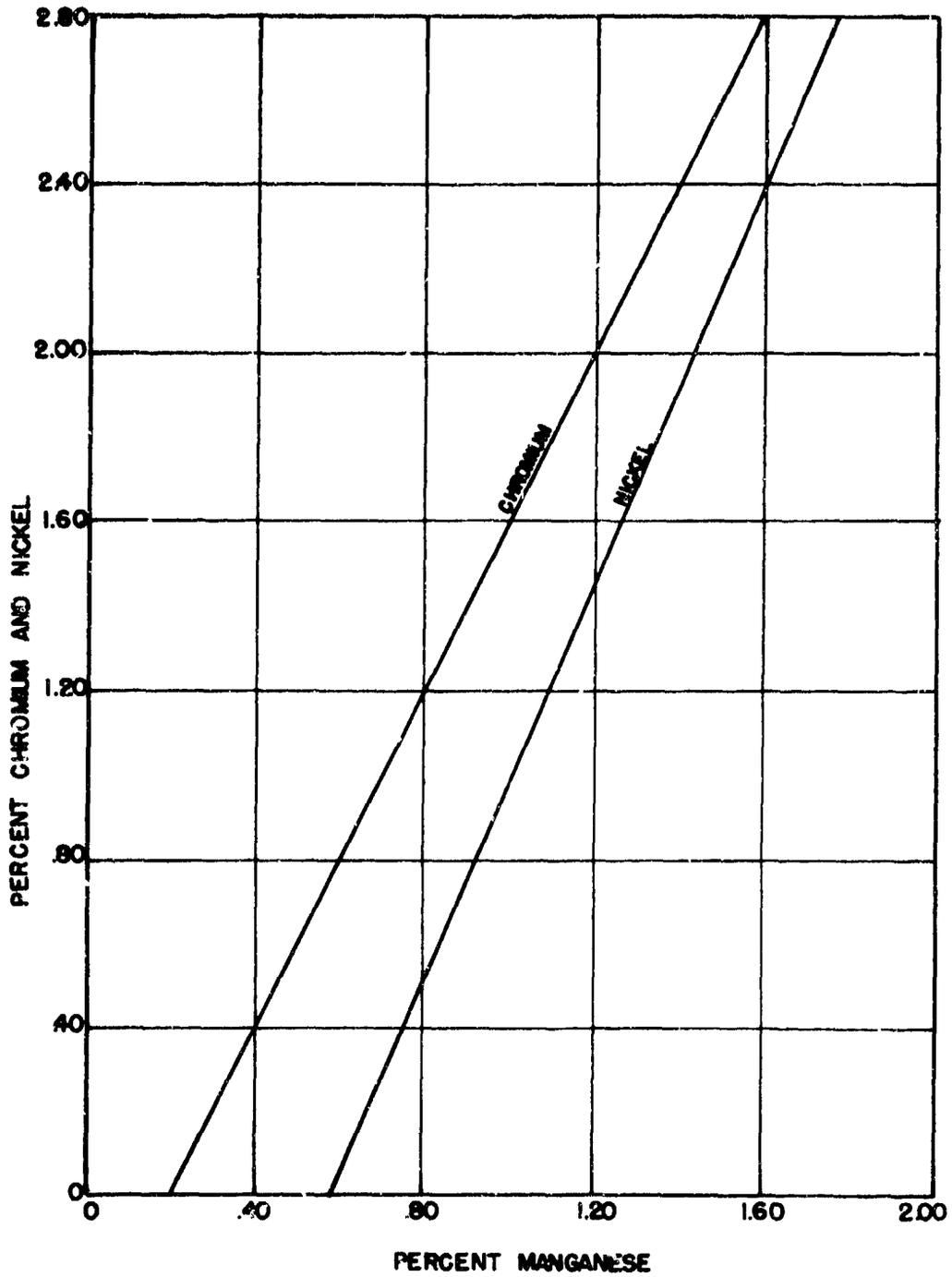

FIGURE 112

OPTIMUM PROPORTIONS FOR MANGANESE, NICKEL AND CHROMIUM BASED UPON MAXIMUM BAINITIC HARDENABILITY WITH MINIMUM LOWERING OF M_s TEMPERATURE. (AFTER JAFFE & HOLLOMON[118])

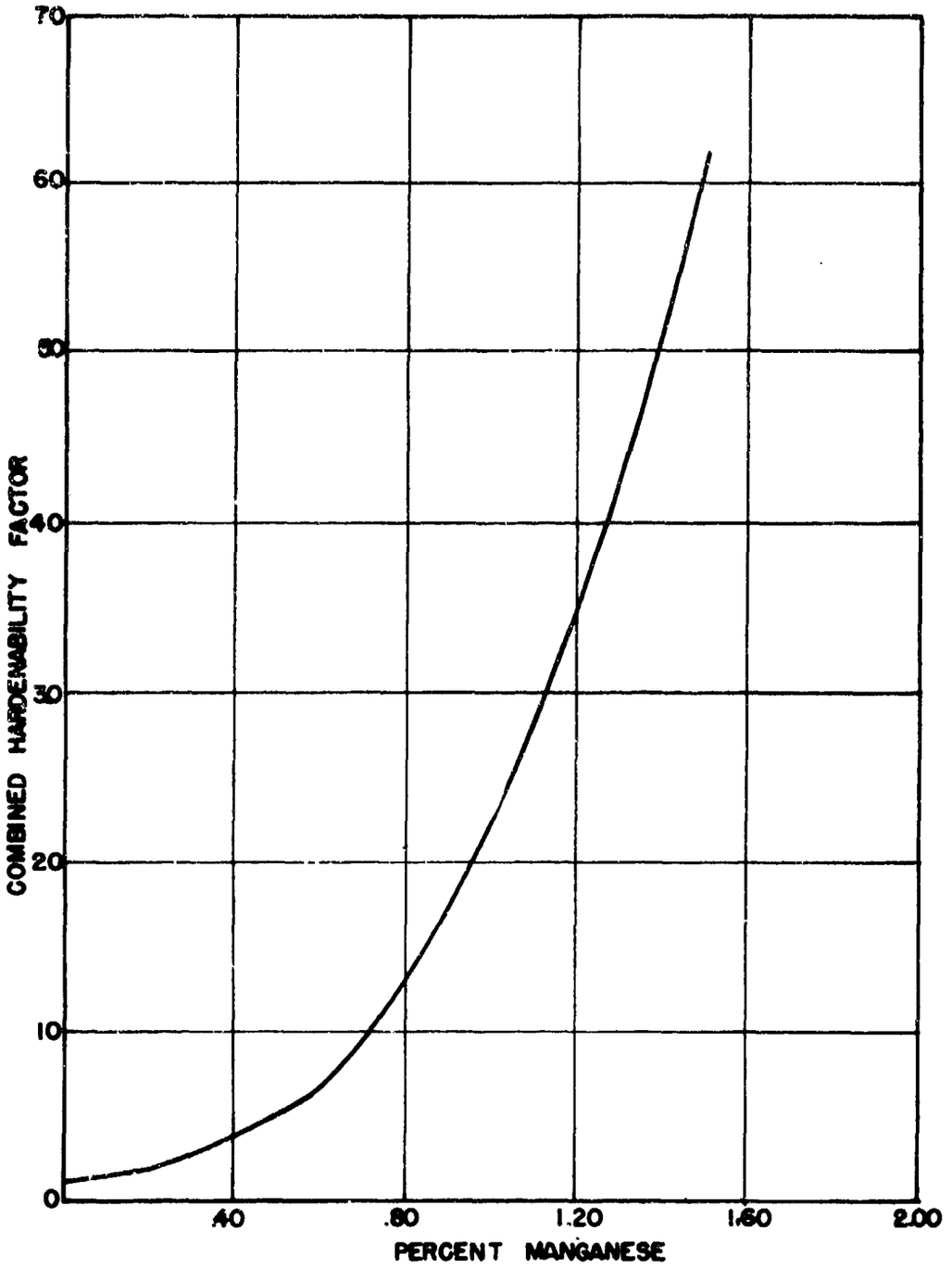

FIGURE 113

COMBINED HARDENABILITY FACTOR FOR MANGANESE, NICKEL AND CHROMIUM IN OPTIMUM PROPORTIONS (AS GIVEN IN FIGURE 112). (AFTER JAFFE & HOLLOMON[118]).

Suggested Readings

H. Scott:
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F. R. Palmer:
"The Relation of Design to Heat-Treatment", Metals Handbook, Am. Soc. Metals, Cleveland, 1939, pp. 214-219.

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VIII. TEMPERABILITY

The maximum hardness obtainable is not often desired in a part, even though maximum hardness is associated with maximum flow and fracture stresses. At maximum hardness the separation between the flow and fracture stresses is small and hence the ductility and resistance to brittle failure are small. A fully martensitic steel has practically maximum hardness in the untempered condition*. A slight tempering, sufficient to cause precipitation of carbide from the martensite, does, it is true, result in a slight increase in hardness, but this increase is so small as to be, ordinarily, negligible. The hardness at room temperature of untempered fully martensitic steel depends significantly only upon the carbon content (except for very low carbon steels, where alloy content has some effect)(295, 296). As Figure 114 indicates, the hardness increases with the carbon content**.

It might be thought that the hardness desired for service could be obtained without tempering by adjusting the carbon content of the steel. However, except where the hardness desired is very high, this procedure would be very difficult to control, for the small variations from the desired carbon content that are inherent in present steel-making practices would produce great variations in hardness (Figure 114). Also, with the low carbon contents involved, it would be necessary, in order to obtain the required hardenability, to use alloy contents so high as to be almost prohibitively expensive (unless the part was very thin or the hardness desired very high). Moreover, if no tempering treatment were used, it would be necessary, in most cases, to put the part into service without relieving the internal stresses resulting from quenching. As will be mentioned, these stresses may have several detrimental effects.

In practice, then, it is ordinarily necessary to use a carbon content that provides, in the untempered martensite, a hardness greater than desired for service, and then to reduce the hardness to the desired value by tempering. In metallurgical design, it is, therefore, frequently necessary to determine the combination of composition and tempering treatment that will produce the desired strength (hardness) with maximum ductility and toughness. The transformations that take place during the tempering of steel were discussed in Chapter I. The strength of a tempered martensitic steel is controlled by the size of the carbide particles, and to a certain extent by their distribution and perhaps their type. These in turn are con-

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*Steels that contain large amounts of retained austenite may be appreciably harder after tempering than before, since tempering induces decomposition of the soft austenite to harder products.

**Reports that the hardness does not increase beyond .70 per cent carbon are undoubtedly due to failure to eliminate retained austenite(297, 298). However, it should be recognized that the Rockwell C hardness scale is very insensitive at the high hardnesses involved, and the change of hardness with carbon content may, therefore, not be very great on this scale.

trolled by the composition and the time and temperature of tempering. For steels that contain only small percentages of carbide-forming elements, the strength decreases continuously with increasing tempering time and temperature. (Figure 115 shows the effect of both time and temperature, while Figures 116 to 118 illustrate the effect of temperature on steels tempered for a fixed time.) Alloying elements such as molybdenum, chromium, and vanadium induce the formation of alloy carbides at high tempering temperatures (or long times). These alloy carbides retard the softening and may cause the hardness to increase with temperature or time over a limited range (Figures 119 and 120).

It has been found that the time and temperature necessary to produce a given hardness (or yield strength, or tensile strength) are related in such a way that the degree of tempering can be expressed by a single parameter relating the effects of time and of temperature(122). Thus the strength of a steel is determined only by a single parameter and is not dependent upon the individual values of the time and temperature. The temperability (response to tempering) of a steel may be expressed in graphs in which some measure of the strength is plotted as a function of this parameter. Two such graphs are presented in Figures 121 and 122. Figure 121 is for a plain carbon steel for which the hardness decreases continuously with increasing temperature and time, while Figure 122 illustrates the temperability of a molybdenum steel exhibiting secondary hardness.

Steels of the same carbon content that are completely martensitic upon quenching and that are not susceptible to temper brittleness will have very nearly the same properties if they have the same hardness (strength). Thus, for such steels, all the properties are specified by the temperability curves. If, however, austenite is retained upon quenching, low-temperature tempering may not result in the same transformations of the austenite as high-temperature tempering. The hardness may not be greatly affected thereby but the impact properties after the two treatments may be widely different.

The relation between time and temperature for the temper brittleness transformation is not the same as for the precipitation and agglomeration of the carbides. Above about 1100° F. (600° C.) the temper brittleness precipitation does not occur, while below 1100° F. (600° C.), its time-temperature-transformation curve has the form of a "C" (Figure 18). Therefore, tempering a susceptible steel at temperatures below 1100° F. (600° C.) for a long time will not result in the same toughness as tempering at high temperatures for a short time, even though the hardnesses may be the same. (As far as can be determined the precipitate that results in temper brittleness does not change the hardness of the steel.)

One other peculiarity in tempering behavior must be noted. As indicated in Figure 123, tempering at moderately low temperatures frequently results in impact properties inferior to those obtained by tempering at lower temperatures, even though the hardness is not as high.

Hardness After Tempering

Relation Between Time and Temperature

Before discussing the effects of carbon and of alloying elements upon the strength or hardness after tempering, it is desirable to reduce the two variables of tempering time and tempering temperature to a single variable. It has been found empirically(122) that, over a wide range of times, temperatures, compositions, and structures before tempering, the hardness depends only upon the tempering parameter:

$$M = T\ (\log t/t_o) = T\ (c + \log t), \qquad (8\text{-}1)$$

and not upon the time and temperature separately. Here, T is the absolute temperature, t the time, and t_o or c, a constant*. Examples are given in Figures 121 and 122. (The relation has not been checked adequately for the very low tempering temperatures and times at which the decomposition of tetragonal martensite is the principal reaction.) The value of the constant is not critical and no effect of alloying elements or of prior structure

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*For cases where the tempering cycle cannot be characterized by a single temperature and a single time, more general expressions for the parameter M must be used. If several tempering treatments, each at a fixed temperature, are successively applied to a piece, the value of the parameter at the end of each treatment may be computed as:

$$T \log \left(\frac{\Delta t}{t_o} + 10^{M_o/T} \right)$$

or

$$T \log \left(\Delta t \cdot 10^{c} + 10^{M_o/T} \right)$$

where Δt is the time spent at temperature T and M_o is the value of the parameter at the start of this time. (M_o may be taken as zero for the first treatment.) For cases in which the temperature is changing during tempering, recourse must be had to the differential equation,

$$dM/dt = \frac{T}{2.303\ t_o\ 10^{M/T}} = \frac{T}{(2.303)\ 10^{(M/T)-c}}$$

upon it has been established. Carbon apparently does have a small effect, as indicated in Figure 112. The constant c varies linearly from $19\frac{1}{2}$ at .30 per cent carbon to 15 at 1.10 per cent carbon (for times expressed in hours). For steels of .20 to .40 per cent carbon, the value of $19\frac{1}{2}$ is sufficiently accurate for practical purposes, and the parameter can be taken as T($19\frac{1}{2}$ + log t), for times expressed in hours. On this basis, Figure 113 has been prepared. All the combinations of tempering time and temperature that fall on the same curve in this figure will produce the same strength after tempering, provided the structure prior to tempering is held constant. As an indication of the effect of carbon content, Figure 125 may be compared with Figure 126, which is drawn for steels of .90 to 1.20 per cent carbon (c equal to 15). The difference between the two figures is not great.

Steels Containing Little or No Carbide-Forming Elements

Alloying elements may affect hardness response to tempering in two ways: primarily, through their effect upon the structure obtained with a given tempering treatment, and secondarily, through their effect upon the hardness of the ferrite in a given metallographic structure. Nevertheless, it has been established that steels containing little or none of the carbide-forming elements have hardness-vs.-parameter curves that are practically linear. This is true up to within 3 points Rockwell C (about 30 points Brinell) of the hardness prior to tempering, and holds on the Rockwell scale down to about C25 and on the Brinell scale to below 175. (See Figure 121.) Moreover, the slopes of these curves are essentially the same, regardless of the composition and the structure before tempering, (provided the same value of c is used in plotting the curves). For steels containing .20 - .40 per cent carbon, the hardness is given approximately, within the limits mentioned, by the equations:

$$\text{Hardness} = H_c - H_d \qquad (8\text{-}2)$$

$$H_d = .00216T(19.5 + \log t) \text{ Rockwell C}$$
$$H_d = .0190T(19.5 + \log t) \text{ Brinell} \qquad (8\text{-}3)$$

Here T is the temperature in degrees Rankine (° F. + 460°) and t is the time in hours. H_d is termed the "hardness difference" and H_c the "characteristic hardness" of the steel. The characteristic hardness depends upon the composition and the structure before tempering, but is not equal to and should not be confused with the hardness before tempering(122).

Hardness difference values, derived from equation (8-2), are given in Figures 125 and 126. The difference between the hardnesses resulting from any two tempering cycles will be given by the difference between the values indicated for the curves upon which the cycles fall. Thus, if the hardness produced by one tempering cycle is known, the hardness that will

be obtained with all other tempering cycles can be readily found*†.

The characteristic hardness is of use primarily for comparing different steels. It provides a single number that defines the entire response of a steel to tempering, for a given structure prior to tempering. (This does not include retained austenite or temper brittleness effects.) Once the characteristic hardness is known, subtracting from it the hardness difference corresponding to any tempering cycle will give the hardness resulting from that cycle**. There remains, however, the problem of determining the characteristic hardness.

The characteristic hardness of a steel, having a given structure prior to tempering, can be obtained if the hardness resulting from a single tempering cycle, starting with this structure, is known. Thus, if a sample of a steel is available, it is merely necessary to austenitize and quench the sample to produce the intended structure prior to tempering, temper it at any convenient temperature for any convenient time, and measure its hardness. The hardness difference corresponding to the tempering treatment used (Figures 125 or 126) is then simply added to the measured hardness. The result is the characteristic hardness (equation (8-2))***. Figures 125 and 126 will then give, at least to a first approximation, the hardness obtainable with any tempering cycle. If no comparisons with other steels are to be made, the hardness obtainable with any tempering cycle can be obtained, by use of the figures, directly from the measured value, and the characteristic hardness need not be considered.

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*For example, suppose a .30 per cent carbon steel (containing little carbide-forming elements) has a hardness of 32 Rockwell C when tempered 1 hour at 900° F. This treatment corresponds (Figure 125) to a hardness difference of 57½ Rockwell C, while 4 hours at 1000° F. corresponds to 63½ Rockwell C. 63½ - 57½ = 6 Rockwell C. A tempering treatment of 4 hours at 1000° F. would, therefore, result in a hardness of 32 - 6 = 26 Rockwell C.

†Hardness difference values may also be used if several tempering treatments are successively applied to the same piece. If a treatment consisting of 2 hours at 900° F. (hardness difference 58½, according to Figure 113) is followed, after cooling, by another treatment of 1 hour at 900° F., the two treatments are equivalent to a single treatment of 3 hours at 900° F. (hardness difference 59). Again, the 2 hour treatment at 900° F. is equivalent (Figure 125) to 11 hours at 850° F. If a 2 hour treatment at 900° F. is followed by 5 hours at 850°, the hardness difference corresponds to 16 hours at 850° F., and is equal to 59 Rockwell C.

**For example, if the characteristic hardness is 95 Rockwell C, a tempering treatment of 1 hour at 900° F., which provides a hardness difference of 57½ (Figure 125), would result in a hardness of 95 - 57½ = 37½ Rockwell C.

***For example, if the tempering treatment of 1 hour at 900° F., corresponding to a hardness difference (Figure 125) of 57½, resulted in a hardness of 37½ Rockwell C, the characteristic hardness would be 57½ + 37½ = 95 Rockwell C.

Unfortunately, practically no work has been done on the relation between characteristic hardness and composition. Reliable data are at present available only for the effect of silicon (in martensitic steel, .50 - .55 per cent carbon) and for the effect of cobalt (in martensitic steel, .40 - .45 per cent carbon). Silicon up to 3.80 per cent increases the characteristic hardness, linearly, about 4 Rockwell C per per cent. (Figure 127.) Cobalt up to 7.40 per cent increases the characteristic hardness, linearly, about 1/2 Rockwell C per per cent. There is no good information available as to the effects of the other elements. A very tentative curve for the effect of carbon upon the characteristic hardness of tempered-martensitic iron-carbon alloys is presented as Figure 128*. In the absence of better data, it may be estimated from Figures 117 to 120 that 1 per cent manganese increases the characteristic hardness 4 Rockwell C, 1 per cent nickel 1 Rockwell C, chromium, in small amounts, 1½ Rockwell C per per cent, and molybdenum, in small amounts, perhaps 2½ Rockwell C per per cent (Table VII). Even the additivity of the effects of the various elements has not been established. Though additivity may be assumed as a first approximation, there are some indications that additions of alloying elements to a highly-alloyed steel do not increase the characteristic hardness as much as the same additions increase the hardness of a steel containing little or no alloy.

When no experimental results are available for calculating the temperability of a steel, Table VII and Figure 128 provide a means of obtaining an estimate of the characteristic hardness from composition. However, because the data upon which the table and figure are based are so scant, any information on the temperability of steels similar to the one whose temperability is sought is likely to prove useful, provided the steels have the same structure prior to tempering. The characteristic hardness of a similar steel can be obtained if the hardness resulting from one tempering cycle is known. The difference between this known characteristic hardness and that of the steel whose temperability is sought can be estimated from Table VII and Figure 128, and the characteristic hardness of the latter steel thus obtained. The hardness that will result from any tempering cycle applied to this steel can then be found by using Figure 125 or 126.

Steels Containing Carbide-Forming Elements**

As has been discussed in Chapter I, when moderate or large amounts of carbide-forming elements are present, a complex alloy-base carbide is likely to precipitate at high tempering temperatures and long tempering times. Such a precipitate appreciably retards the softening of the steel. (This phenomenon is known as "secondary hardening".) Therefore, although the same tempering parameter and tempering constant seem to apply whether or not car-

*This curve was obtained from a published curve(122) for plain-carbon steels by subtracting hardness increments corresponding to the manganese and silicon contents of the steels.

**Most of the conclusions of the following paragraphs are based on an earlier study(299).

bide-forming elements are present, the linear relationship between hardness and tempering parameter does not hold when moderate or large percentages of these elements are present. Superimposed on the normal softening associated primarily with the growth of cementite particles is an age-hardening, associated with the nucleation and growth of alloy carbides, and entirely similar to the age-hardening phenomena well-known in non-ferrous alloys.

In steels containing more than .25 per cent molybdenum, for example, the normal linear behavior holds up to tempering treatments of about 1 hour at 850° F. (450° C.), or any equivalent combination of time and temperature. On further tempering, the hardness falls off less rapidly, and if the molybdenum and carbon contents are high enough it may actually increase. (Figures 119 and 122.) The secondary hardness reaches a maximum after a tempering treatment equivalent to about 1 hour at 1100° F. (600° C.). As tempering is continued, the hardness falls off as rapidly as it does for plain-carbon steels, or even somewhat more rapidly.

Vanadium steels behave similarly to molybdenum steels, but less of the alloying element is needed to produce the same amount of secondary hardening. The effect of tungsten is probably similar, but it might be expected that (because of the high atomic weight of tungsten) more tungsten than molybdenum would be required to produce a given degree of secondary hardening.

The effects of chromium, though qualitatively rather similar to those of molybdenum, manifest themselves at somewhat lower values of the tempering parameter. (Figure 120.) Apparently the chromium carbide precipitate forms sooner and grows somewhat more rapidly than the molybdenum carbide. Secondary hardening in high-chromium steels begins to be noticeable at approximately 1 hour at 650° F. (350° C.), and reaches its maximum at perhaps 1 hour at 850° F. (450° C.), where the effect of molybdenum first appears.

For the prediction of temperability from composition, the assumption of a linear relationship between hardness and the parameter M may be used for tempering treatments (values of M) up to those at which secondary hardening begins. For tempering treatments (values of M) greater than those necessary to develop full secondary hardening, the same assumption can be made. The decrease in hardness with increasing M is actually somewhat greater after full secondary hardening has occurred (while "over-aging" is taking place) than if there had been no secondary hardening, but to a rough approximation, it can be considered the same and the hardness difference values of Figure 125 or 126 employed. The characteristic hardness is, however, considerably different after secondary hardening than before. The range of tempering treatments in which secondary hardening is increasing can be considered as one in which the characteristic hardness is increasing.

The characteristic hardnesses, both before and after secondary hardening, of the chromium steels and molybdenum steels covered by Figures 119 and 120 are shown in Figures 129 and 130. Corresponding values for the effect of moderate percentages of chromium and molybdenum upon the characteristic hardness are given in Table VII. The characteristic hardness of series

of chromium-molybdenum steels are indicated in Figures 131 and 132. It will be seen that the presence of more than .50 per cent molybdenum markedly decreases the effect of chromium upon the characteristic hardness after secondary hardening. It may be surmised that the molybdenum combines with the carbon to form a molybdenum-base carbide, and does not leave enough carbon for an appreciable amount of chromium-base carbide to form. However, even when 5 per cent molybdenum is present, secondary hardening commences at the low values of the tempering parameter characteristic of chromium-carbide precipitation. It is possible that chromium carbide precipitates first and then redissolves as the molybdenum carbide precipitates. In any case, the data indicate that the effects of large amounts of chromium and molybdenum upon the hardness are not additive after secondary hardening has taken place.

Figure 133 illustrates the effect of vanadium upon the characteristic hardness of a steel containing 1 per cent chromium. It is of interest to note that while .16 per cent vanadium does not affect the hardness prior to secondary hardening, it greatly increases the hardness after secondary hardening.

Because of the complexities involved in calculating tempered hardness from composition when secondary hardening occurs, it is advantageous to use whatever information is available on the tempering of steels similar to the one whose tempering behavior is sought. Curves similar to Figures 116 to 120 can be helpful. For such comparative data to be of much use, it must apply to the same structure prior to tempering. In this connection, it must be remembered that it is primarily the composition of the martensite, and not the overall composition of the steel, that determines the temperability. For example, large particles of alloy carbide, as are likely to result from insufficient austenitizing, do not have much effect upon temperability and the carbon and alloying elements combined in them should be deducted from the overall composition before evaluating temperability.

Brittleness on Low Temperature Tempering

As Figure 123 indicates, tempering treatments equivalent to about 1 hour at 450-650° F. (250-350° C.) generally seem to result in impact properties inferior to those obtained with less tempering, even though the hardness is lower. The phenomenon has not been investigated in detail, but changes in carbon and alloy content have not yet been found to affect it. It may be hypothesized that as carbide particles precipitate from the martensite on tempering and first begin to grow, their shape, size, and distribution are such that for a while the fracture strength of the steel falls more rapidly than the flow stress. This necessarily reduces the energy necessary to fracture the steel under complex stresses and high rates of strain. (Chapters III and IV.)

Temper Brittleness

The properties of steels susceptible to temper brittleness may be affected by the temper brittleness precipitation, which tends to occur in

such steels at temperatures below 1100° F. (600° C.). As discussed earlier, the precipitate lowers the fracture-stress curves and so changes the impact properties, but does not affect the flow-stress curves (or hardness). The effect of the precipitate is to shift the notched-bar impact curve, raising the temperature of brittle failure. Thus, the apparent effect of the precipitate at any one temperature of notched-bar testing not only depends upon the amount of the precipitation that has occurred but also upon the position of the notched-bar impact curve of the unembrittled steel. (See Figure 79.) Since no direct measure of the amount of precipitation has been devised and since it is difficult to obtain quantitative measures from the notched-bar data even when the entire notched-bar energy curves are available, the quantitative effects of changes in composition on the rate and extent of the precipitation have yet to be determined. The available data indicate that the three elements manganese, nickel, and chromium increase the amount of precipitation that occurs with a given treatment below 1100° F. (600° C.), while molybdenum, at least up to .40%, decreases the amount (for short and moderate tempering times). Based upon notched-bar data for tempered martensitic steels, it appears that the relative effects of the three alloying elements in increasing the susceptibility are similar to their relative effects in increasing the hardenability. Thus, manganese has a large effect on hardenability (both bainitic and pearlitic) and also markedly increases the susceptibility toward temper brittleness. Nickel and chromium have less effect than manganese both on hardenability and on temper brittleness. In Figure 134, a relative measured susceptibility to temper brittleness is plotted as a function of the combined pearlitic hardenability factors for nickel, chromium, and manganese*. The higher the molybdenum content the higher can the manganese, nickel, and chromium contents be raised without producing embrittlement (with a fixed treatment for moderate times). It is not known definitely whether carbon, phosphorus, sulphur, silicon, or any of the other elements that may be present in steel affect the susceptibility(129).

For steels susceptible to temper embrittlement, it is often desirable to temper to a given hardness in such a fashion as to minimize the amount of temper brittleness precipitation that will occur. For moderately susceptible steels, at least, tempering above the temperature at which the precipitation occurs, followed by rapid cooling, will prevent the precipitation. An indication of the effect of cooling rate upon the development of temper brittleness is given in Figure 135. It appears that, in general, the less susceptible the steel is, the more slowly it may be cooled following tempering, without danger of embrittlement. For heavy sections it is usually necessary to use very susceptible steels (to obtain the hardenability) and the cooling rates are so very slow that embrittlement may occur even upon water quenching the part from above 1100° F. (600° C.).

In many cases the tempering time can be made sufficiently short to permit tempering above 1100° F. (600° C.). In other cases it cannot. A typical example of the variation of impact strength obtained (without change in

*The steels for which the data were obtained contained about .30% carbon and from .25 to .40% molybdenum(129).

hardness), by holding a steel for various times and temperatures below 1100° F. (600° C.) is given in Figure 136. The C-shape of the curve is to be noted, as well as the fact that the eventual embrittlement is greater the lower the temperature. Since above the nose of the C curve (975° F. (520° C.)) the time necessary to begin the precipitation increases with temperature, tempering at high temperatures (above 975° F.) will be less likely to produce embrittlement. Since at low temperatures the time necessary to obtain a given hardness is more dependent upon temperature than is the time necessary to obtain a given amount of precipitation, tempering at the highest possible temperature and shortest time will always produce the minimum embrittlement at a given hardness (Figure 139). The highest possible temperature is ordinarily set by the shortest tempering time that can be used, through the time-temperature relation for the desired hardness. If the tempering temperature is above the nose of the temper brittleness C curve, rapid cooling following the temper will help to minimize temper brittleness.

Retained Austenite

As has been mentioned, retained austenite is likely to decompose on tempering to bainite or pearlite, which are undesirable in steels intended to be fully hardened. The amount of austenite retained is decreased by decreasing the temperature to which the steel is cooled after quenching. Also, the less the time consumed, within the martensite range, in bringing the steel to this minimum temperature, the less the austenite that is retained.

It is not always practical to cool a steel part below the M_f temperature, particularly if this lies below room temperature. For instance, suitable cooling equipment is not generally available for very large parts. Moreover, as indicated in Chapter I, even cooling without delay to very low temperatures does not always result in complete transformation of austenite to martensite. When it is not practical to transform all the austenite by cooling, the best procedure that can be followed is, ordinarily, to transform the austenite to the product that has the least detrimental effect upon the properties. This product is lower bainite. (In those cases where no tempering is necessary, the best procedure, if dimensional stability is not essential, may be to leave the austenite untransformed.) The austenite may be transformed to lower bainite by a preliminary tempering treatment in the lower portion of the bainite range, followed by normal tempering to produce the desired hardness. Since the presence of martensite accelerates the transformation of austenite to bainite, only a brief preliminary temper is usually necessary (at least for hypoeutectoid steels). In most cases, gradual heating to the normal tempering temperature, as occurs when the part is heated in a mild heating medium such as air, is sufficient to decompose the austenite to lower bainite.

Residual Stresses

Residual stresses may be introduced by transformations and by temperature differences within the part during quenching or other heat treatment, as has been mentioned in Chapter VII. They may also be introduced by

direct plastic deformation of the part (Chapter III). If the plastic flow or transformation is non-uniform on a microscopic scale, the residual stresses will be on the same scale and are often referred to as tessellated, parasitic, or Heyn stresses. If the non-uniformity is on a larger scale, the residual stresses will be on a larger scale.

The parasitic stresses give rise to the Bauschinger effect, as has been discussed in Chapter III. Deviation of the stress-strain curve from linearity may also be caused by parasitic stresses introduced by, for example, quenching. It is believed by many that parasitic stresses arising from quenching and transformation serve to decrease the ductility at high hardnesses. A mechanism has been advanced(103) to predict such an effect but confirmatory experiments are difficult to design. Probably the most important effect of these strains is to lower the endurance limit in fatigue.

Macroscopic residual stresses are extremely important, for they cause "warping" or distortion upon heating or machining. Favorable residual stresses may be intentionally introduced into metal parts to increase the external loads, in certain directions, necessary to cause yielding or fracture. Autofrettage of gun tubes and shot peening of springs are excellent examples of the introduction of "favorable" residual macrostresses. Unfavorable residual macrostresses decrease the overall strength of the metal part because they lower the external load necessary for yielding. Residual stresses in steels subject to brittle failure may increase the temperature at which such failure occurs, because they constrain the metal and raise the flow stress curve. Further, if the stresses are "unfavorable", brittle failure may occur at smaller values of the applied load.

Frequently it is desired to remove or reduce the residual stresses. Parasitic stresses are relieved in order to straighten the initial part of the stress-strain curve and to prevent plastic deformation and fatigue failure at small values of the applied load. Macroscopic stresses are relieved primarily to obtain dimensional stability in machining or in service, or to reduce stresses that would be unfavorable in service. Many believe that residual stresses are primarily the cause of brittle failure and relieve stresses simply to prevent such failures. As pointed out earlier, however, if the metal is ductile, residual stresses generally will not cause brittle failure. It is only that in their presence inherently brittle materials can be made to fail brittlely under less severe conditions of external loading.

Parasitic stresses are reduced by holding at about 600° F. (300° C.) for 1 hour to such an extent that the initial part of the stress-strain curve is raised(302). The Bauschinger effect is also removed by these low temperature treatments. The variation of the yield strength at .01% offset of quenched specimens, tempered for 1 hour at various temperatures, is illustrated in Figure 138. The figure indicates that for this steel the curvature of the initial part of the stress-strain curve is removed at a temperature of about 1100° F. (600° C.). There is little information available regarding the equivalence of the effects of time and of temperature in relieving the parasitic stresses.

Macroscopic stresses may be relieved by two methods:

(1) Removal of stressed metal, which results in a rebalancing of the remaining stresses by deformation (the cause of machining instability),

(2) Plastic deformation, either by applying external loads or by reducing the strength through raising the temperature.

Stresses are generally relieved simply by heating the metal part. Even though many experiments have been performed to measure the variation of stress-relief with time and temperature, difficulties inherent in the experimentation cause the results of such measurements to be very hard to interpret. A recent summary(304) of the available data indicates that the averaged curves of Figure 139 are the best that can be obtained. The scatter in the measurements was so great that no conclusions could be drawn as to the effects of steel composition or of structure on the rate of stress-relief. Figure 139 indicates that moderate changes in stress-relieving time are equivalent to only small changes in temperature. No definite relations have been established between the effects of time and of temperature. How much stress-relief is necessary for a given part must generally be determined by experiment. Stress-relieving and sequence of operations may be adjusted to produce the desired stability. If the macroscopic stresses are unfavorable, it may be necessary to relieve these stresses almost completely to prevent yielding in service. To avoid changing the hardness, stress-relieving must be carried out at temperatures lower than the prior temperature of tempering.

Steels susceptible to temper brittleness may become embrittled upon stress-relieving. If a steel has been quenched after tempering above 1100° F. (600° C.), even if followed by water quenching, stress-relieving at temperatures below 1100° F. (600° C.) may induce the precipitation that results in temper brittleness. Slow cooling from higher stress-relieving temperatures is also likely to embrittle the steel. Rapid cooling will induce additional residual stresses, though they may be favorable. Thus, stress-relieving treatments for susceptible steels must be chosen with full consideration of the temper brittleness transformation.

Removal of Hydrogen

That the presence of hydrogen dissolved in the steel may lower its ductility has already been mentioned. Such hydrogen may be substantially removed by tempering treatments. The higher the temperature (below that at which austenite forms) and the longer the times, the greater the amount removed. The times for removal have been reported(171) to be of the order of 15 minutes at 1300° F. (700° C.), 3 hours at 750° F. (400° C.), and 24 hours at 200° F. (100° C.), for 1/2 inch round bars*. Thus, the time necessary to remove a given amount of hydrogen varies less with temperature than does the time to produce a given hardness (or even the time to produce a given amount

*The time required increases with the thickness, and probably varies directly as the square of the thickness.

of temper brittleness precipitate). If the tempering treatment selected on the basis of hardness and temper brittleness is insufficient for adequate removal of hydrogen and hydrogen embrittlement is a problem, an additional low-temperature, long-time treatment may be used. It should be remembered that removing the hydrogen will not remove flakes that may have been caused by it.

TABLE VII

ESTIMATED EFFECT OF ALLOYING ELEMENTS UPON CHARACTERISTIC HARDNESS OF TEMPERED MARTENSITE*

Element	Rockwell C Increase Per % of Element: No Secondary Hardening	Rockwell C Increase Per % of Element: Full Secondary Hardening
Mn	4	4
Si	4	4
Cr	1-1/2	5
Ni	1	1
Mo	2-1/2	20
V	0	50
Co**	1/2	1/2

*Estimated from data of Bain(299) (except cobalt).
**Estimated from data of Loria(301).

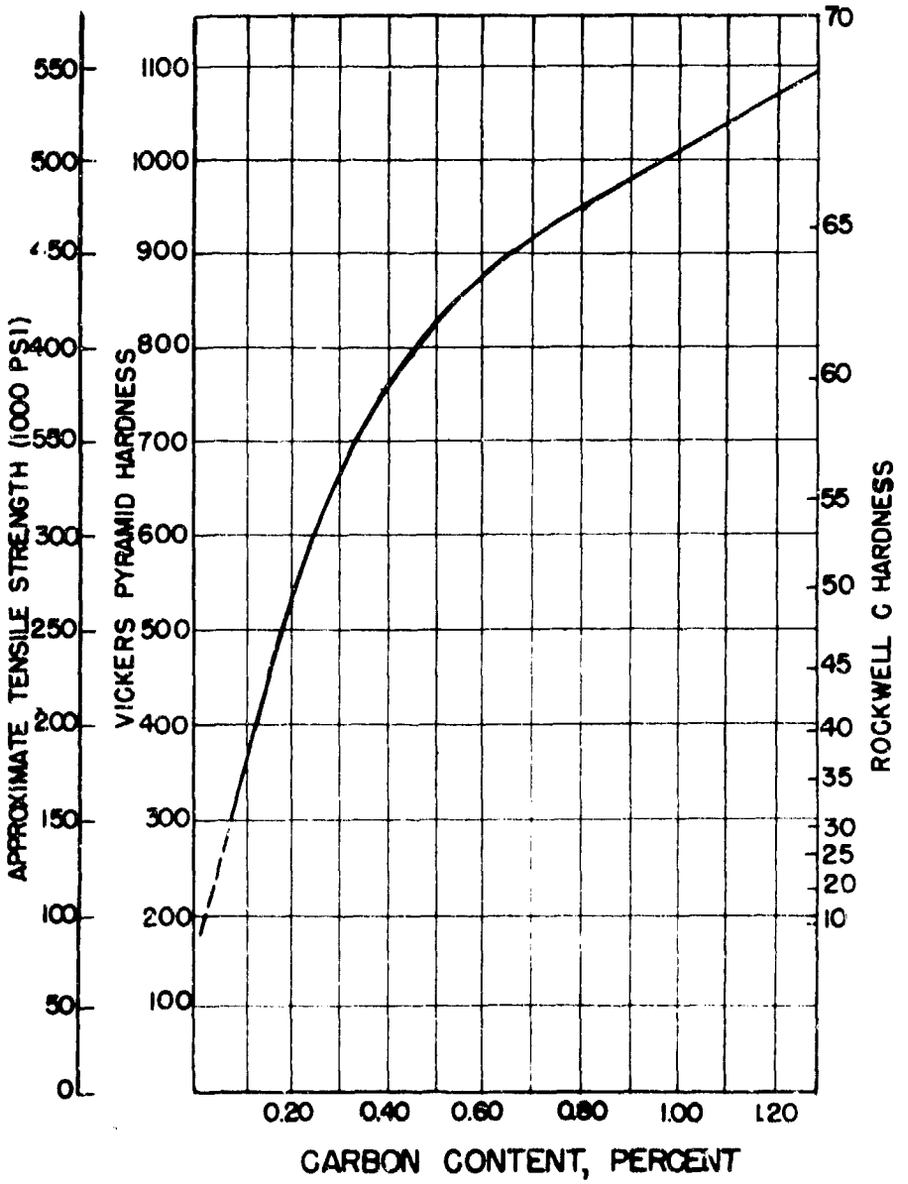


FIGURE 114

EFFECT OF CARBON UPON THE HARDNESS OF FULLY HARDENED STEEL.

(AFTER BAIN,[297] TENSILE STRENGTH SCALE ADDED)

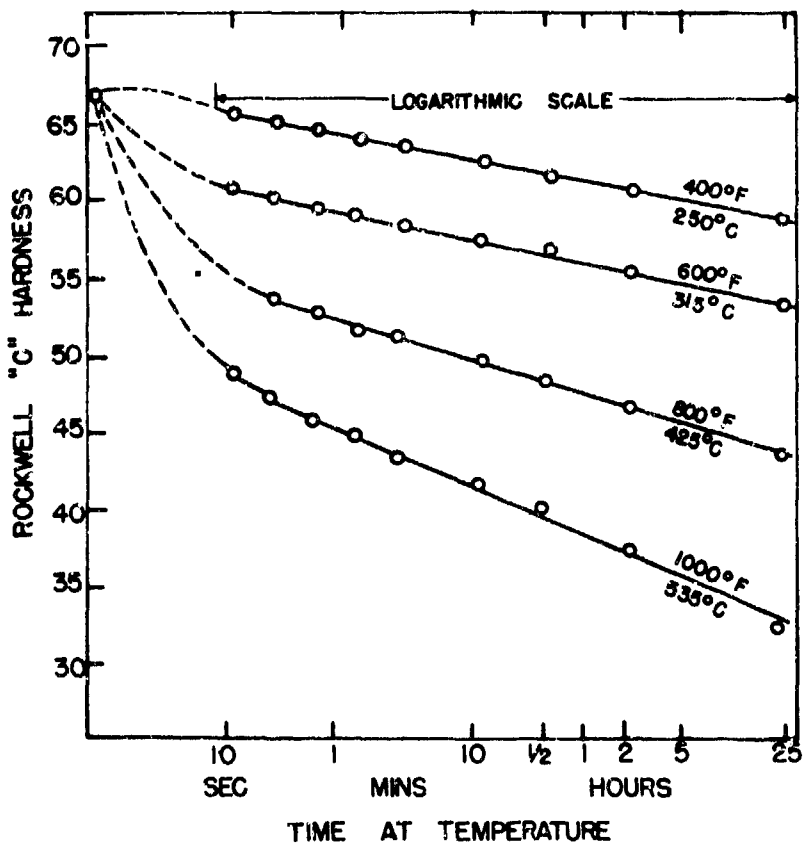


FIGURE 115

EFFECT OF TIME AT SEVERAL TEMPERATURES ON THE HARDNESS OF A .82% CARBON-.75% MANGANESE STEEL. (AFTER BAIN[299])

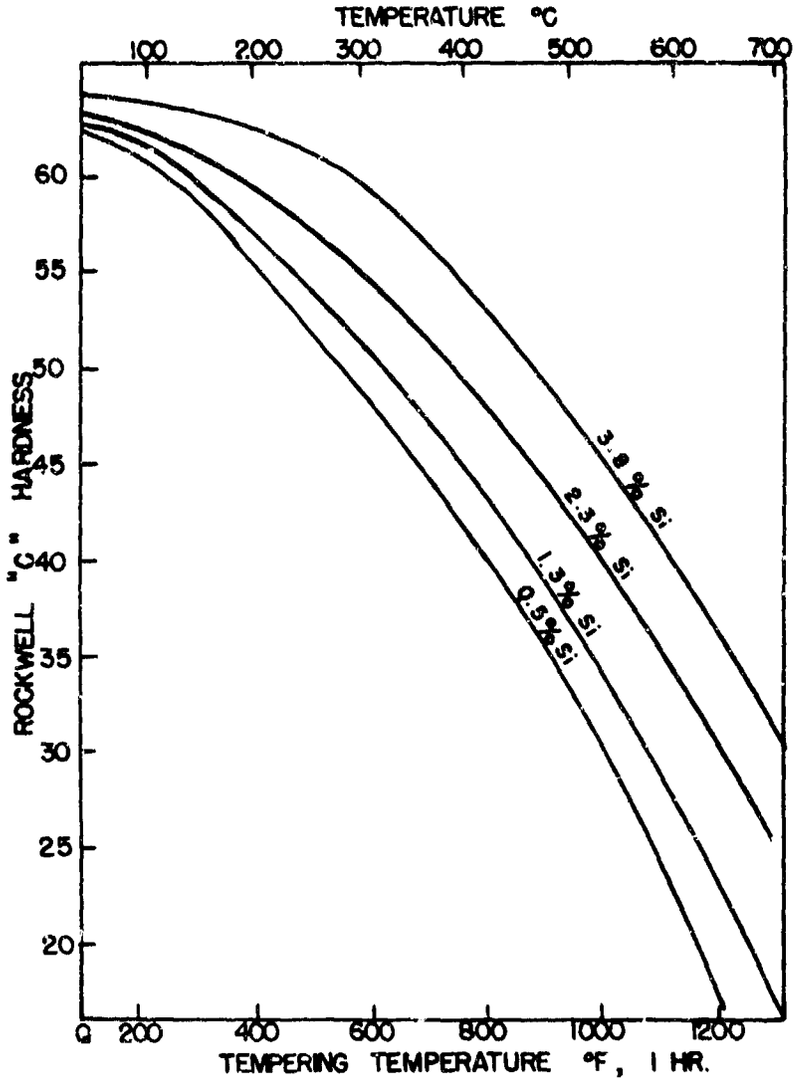


FIGURE 116

EFFECT OF SILICON ON THE TEMPERED HARDNESS OF .50 TO .55% CARBON STEEL. (AFTER BAIN[28])

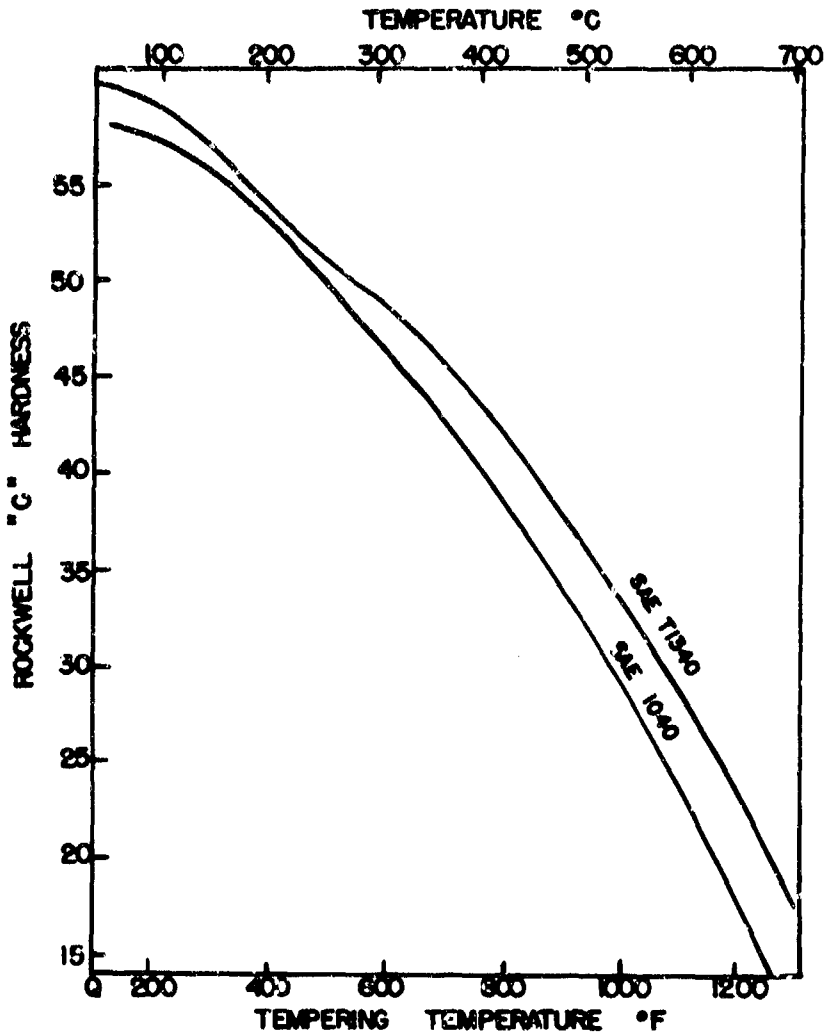


FIGURE 117

EFFECT OF MANGANESE ON THE TEMPERED HARDNESS OF .40 TO .45% CARBON STEEL. (AFTER BAIN[100])

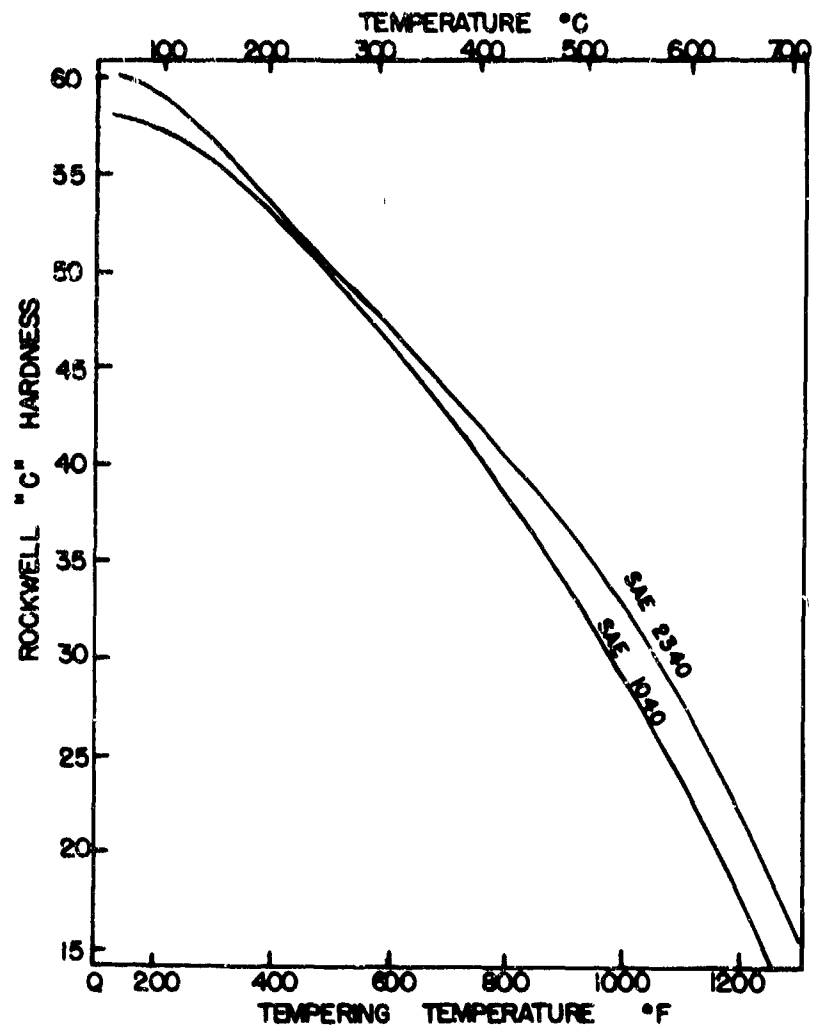


FIGURE 118

EFFECT OF NICKEL ON THE TEMPERED HARDNESS OF .40 TO .45% CARBON STEEL. (AFTER BAIN[28])

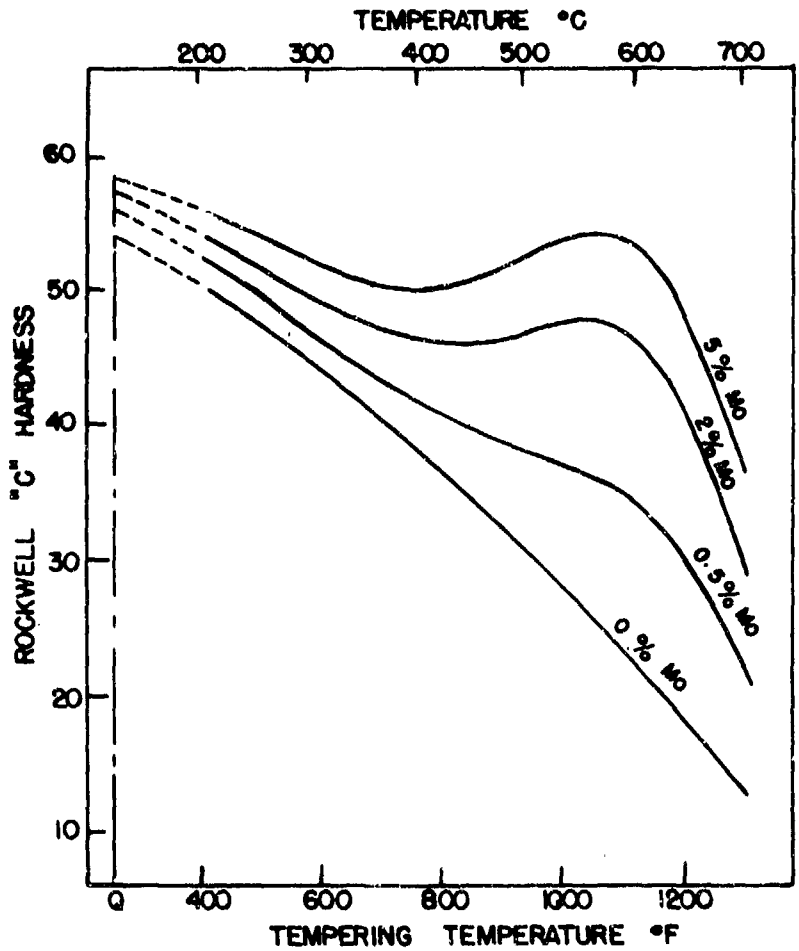


FIGURE 119

EFFECT OF MOLYBDENUM ON THE TEMPERED HARDNESS OF .35% CARBON STEEL. (AFTER BAIN[288])

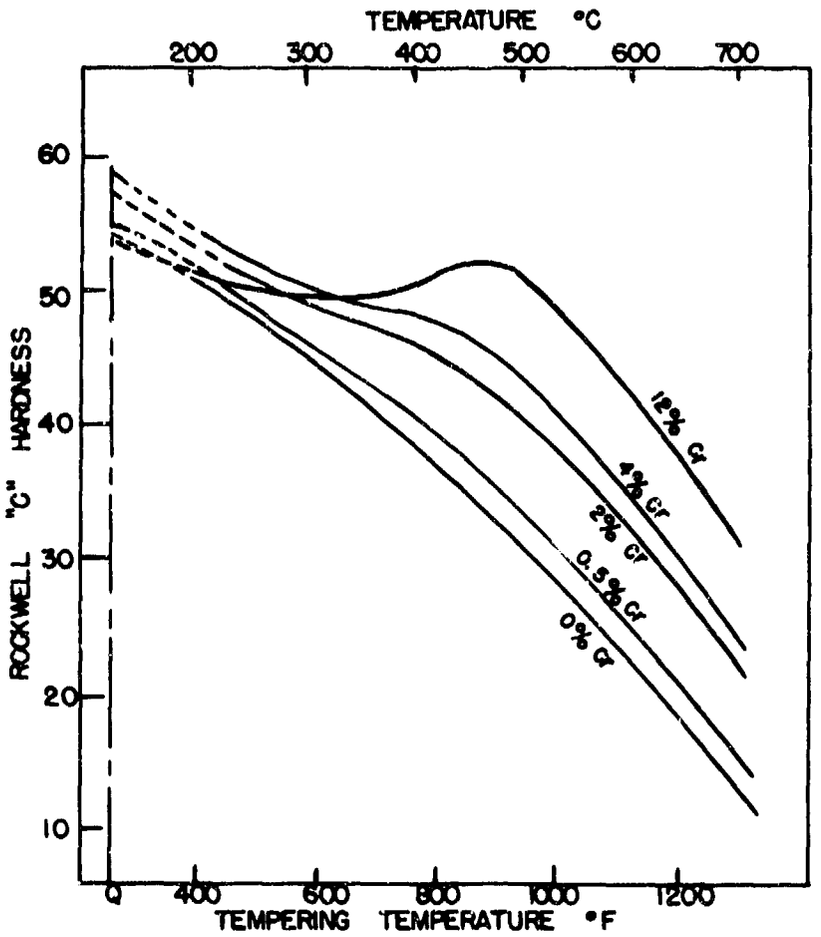


FIGURE 120

EFFECT OF CHROMIUM ON THE TEMPERED HARDNESS OF .35% CARBON STEEL. (AFTER BAIN[256])

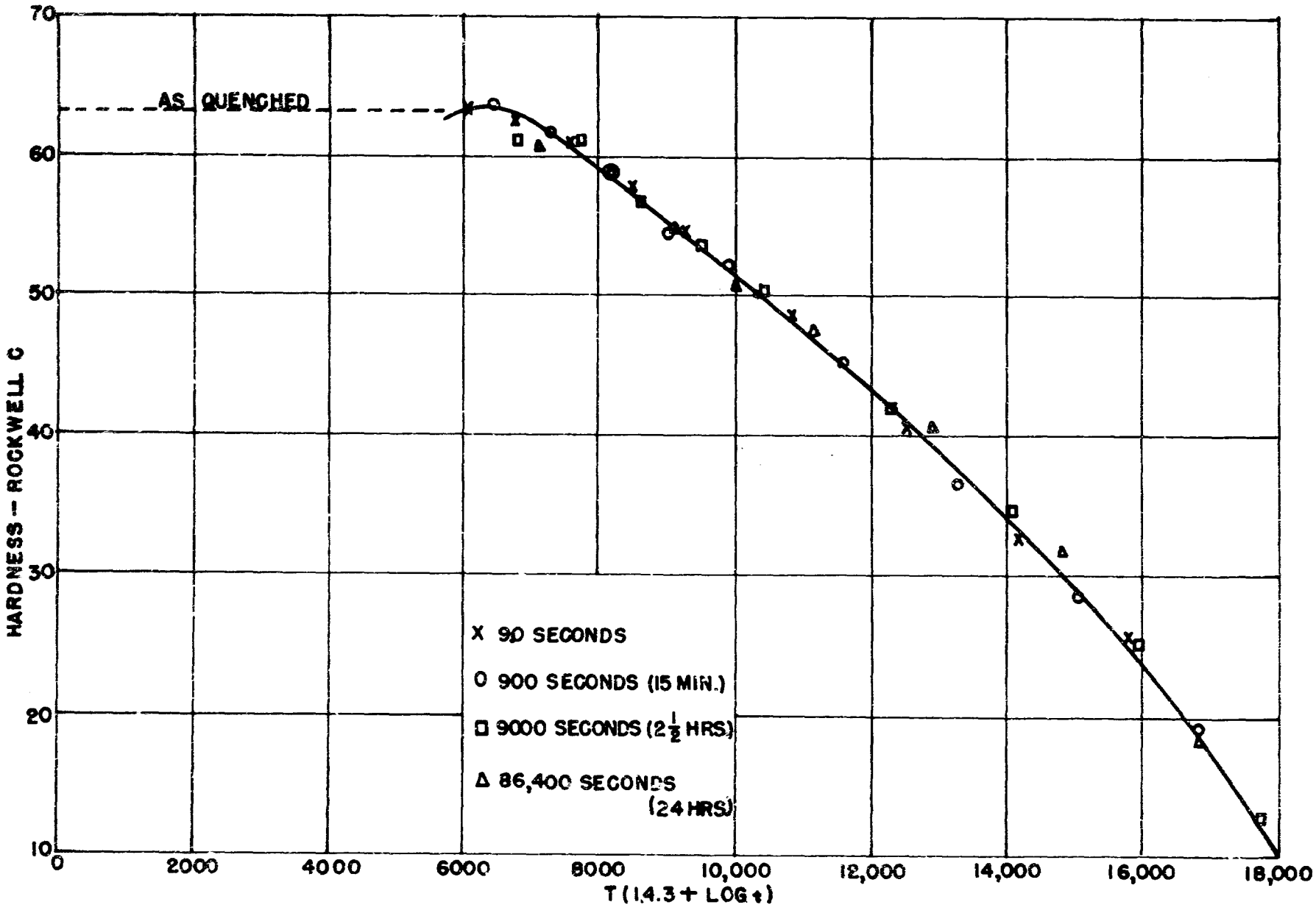


FIGURE 121

TEMPERABILITY OF A PLAIN CARBON STEEL. (AFTER HOLLOMON & JAFFE[122])

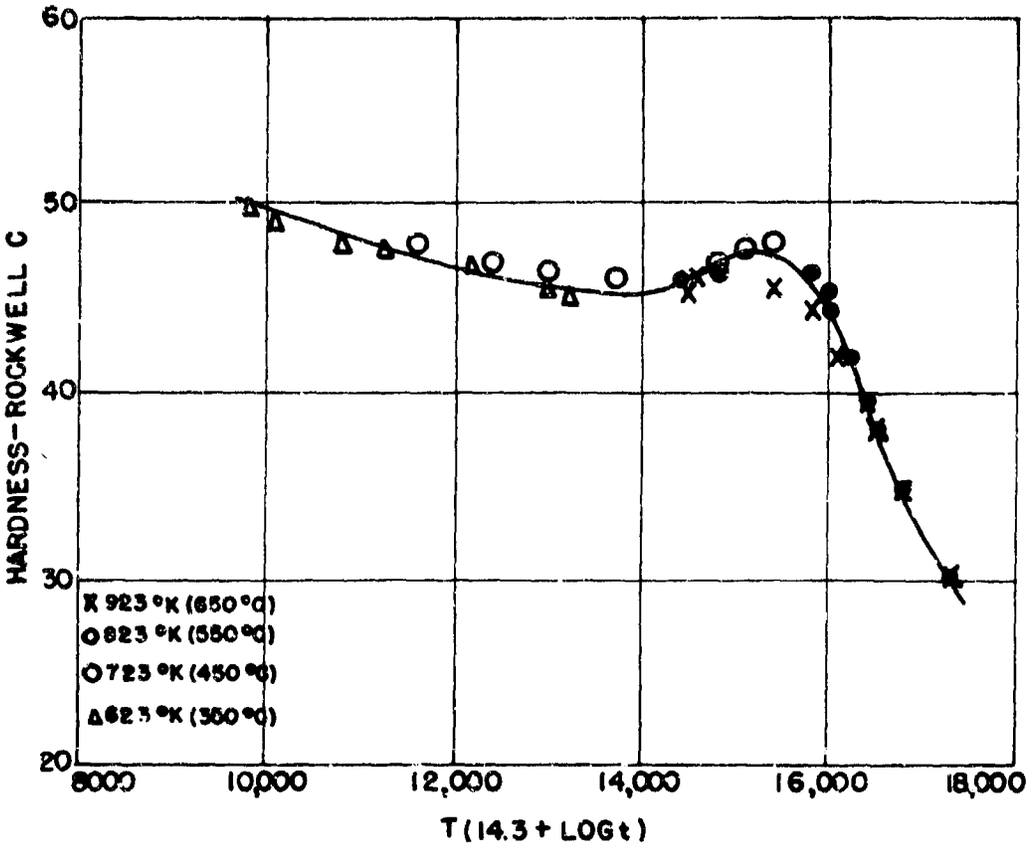


FIGURE 122

TEMPERABILITY OF A STEEL EXHIBITING SECONDARY HARDENING. (AFTER HOLLOMON & JAFFE[122])

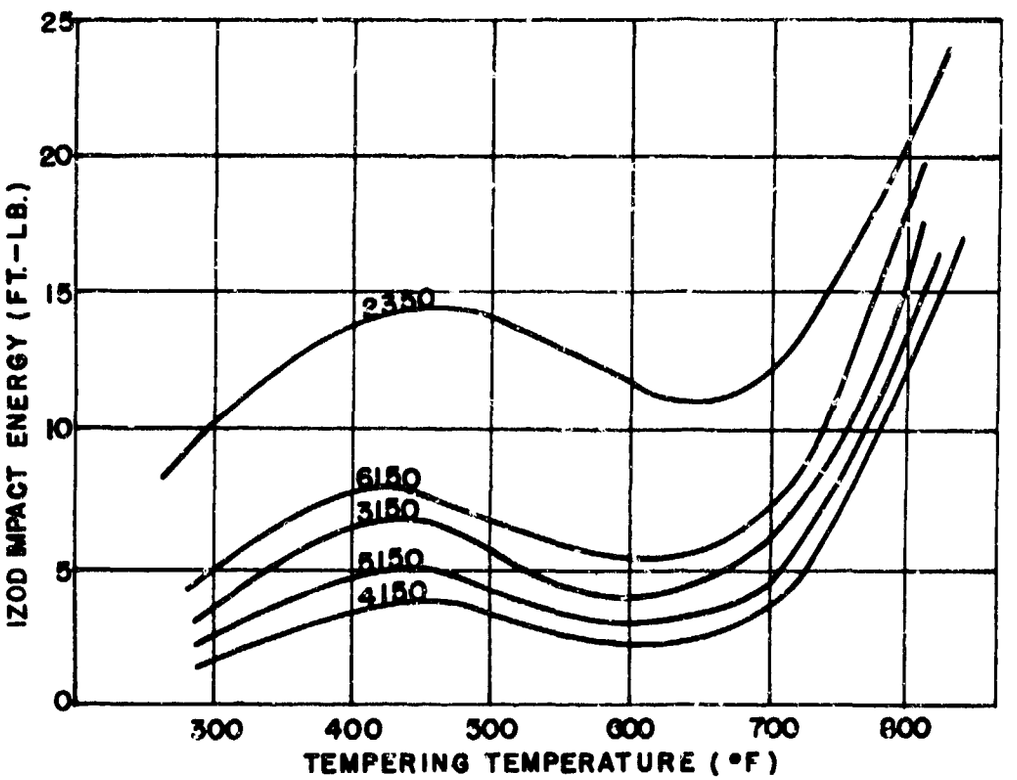


FIGURE 123

EFFECT OF TEMPERING AT LOW TEMPERATURE UPON IMPACT ENERGY FOR VARIOUS FULLY HARDENED SAE STEELS.

(AFTER GROSSMANN[300]).

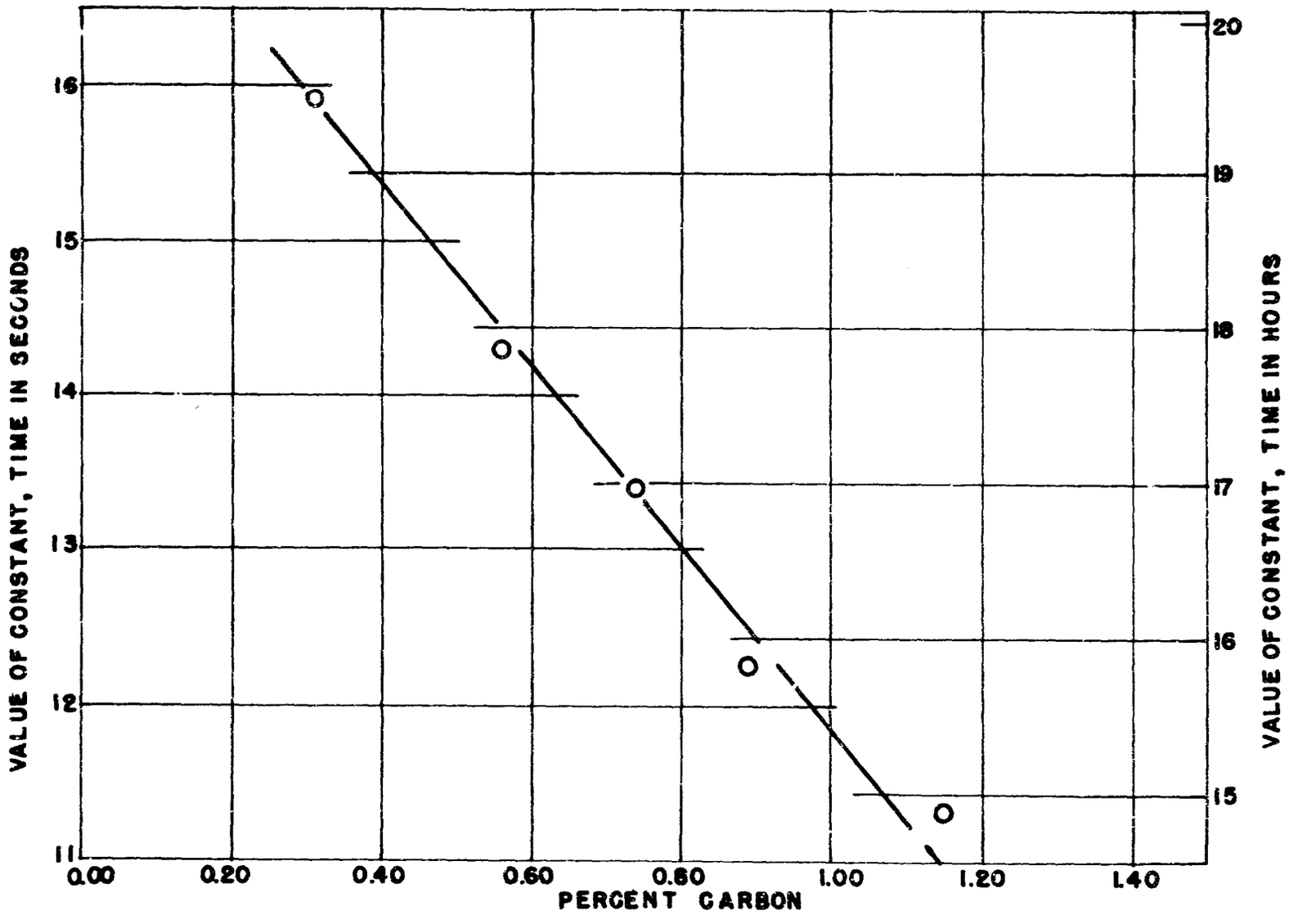


FIGURE 124

VARIATION OF CONSTANT IN TIME-TEMPERATURE PARAMETER WITH CARBON CONTENT FOR TEMPERING FULLY QUENCHED PLAIN CARBON STEELS

FIGURE 125

TIME-TEMPERATURE COMBINATIONS GIVING

EQUIVALENT TEMPERING

PRIMARILY FOR STEELS CONTAINING 0.20-0.40 PER CENT CARBON.

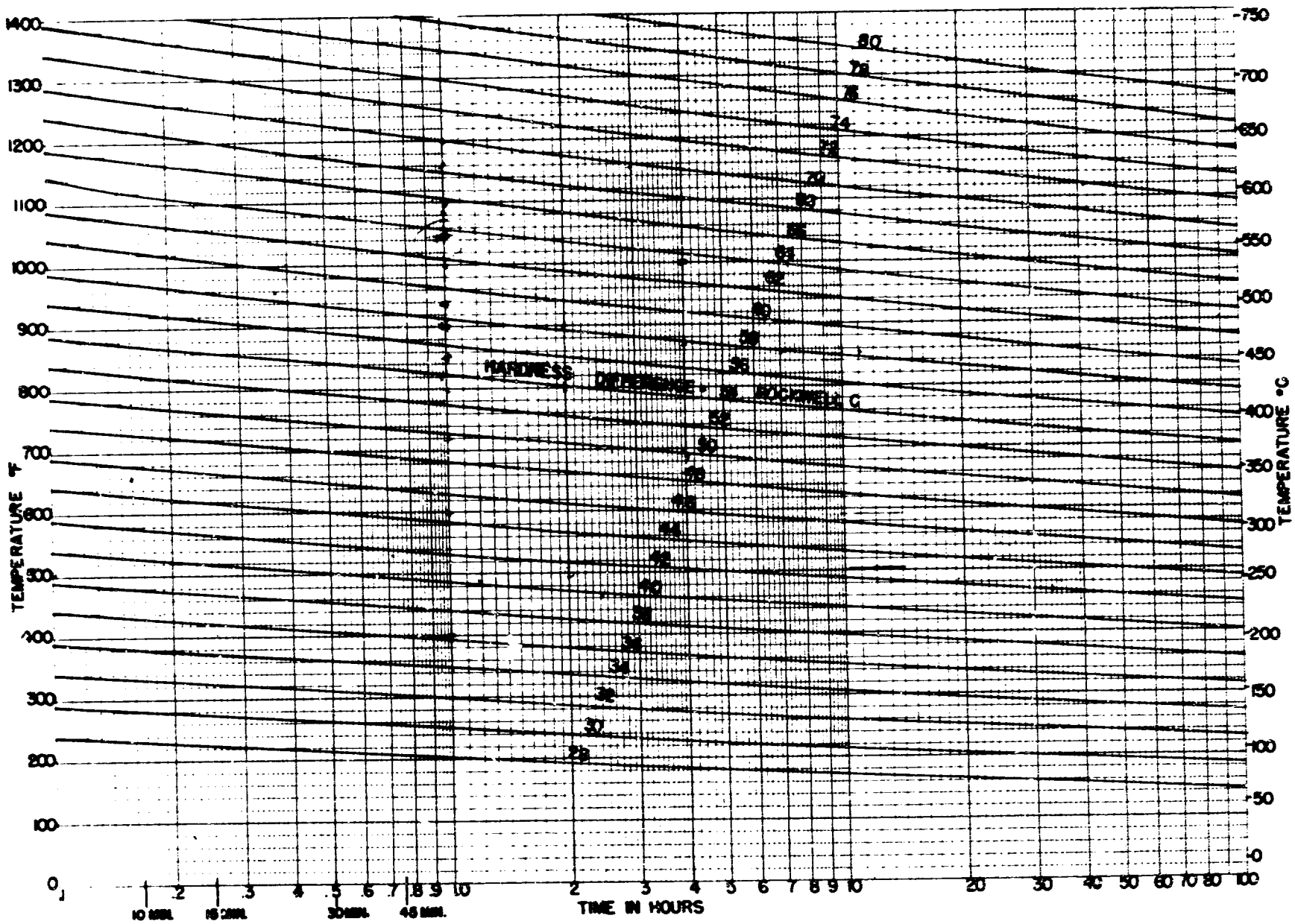
APPROXIMATE HARDNESS DIFFERENCES FOR STEELS

NOT CONTAINING LARGE PERCENTAGES OF

CARBIDE-FORMING ELEMENTS

ARE ALSO SHOWN.

$c = 19\frac{1}{2}$ (TIME IN HOURS).



TEMPERATURE °F
TEMPERATURE °C
TIME IN HOURS
HARDNESS DIFFERENCE - 58 ROCKWELL C
10 MIN.
15 MIN.
30 MIN.
45 MIN.

FIGURE 126

TIME-TEMPERATURE COMBINATIONS GIVING

EQUIVALENT TEMPERING

PRIMARILY FOR STEELS CONTAINING 0.90-1.20 PER CENT CARBON.

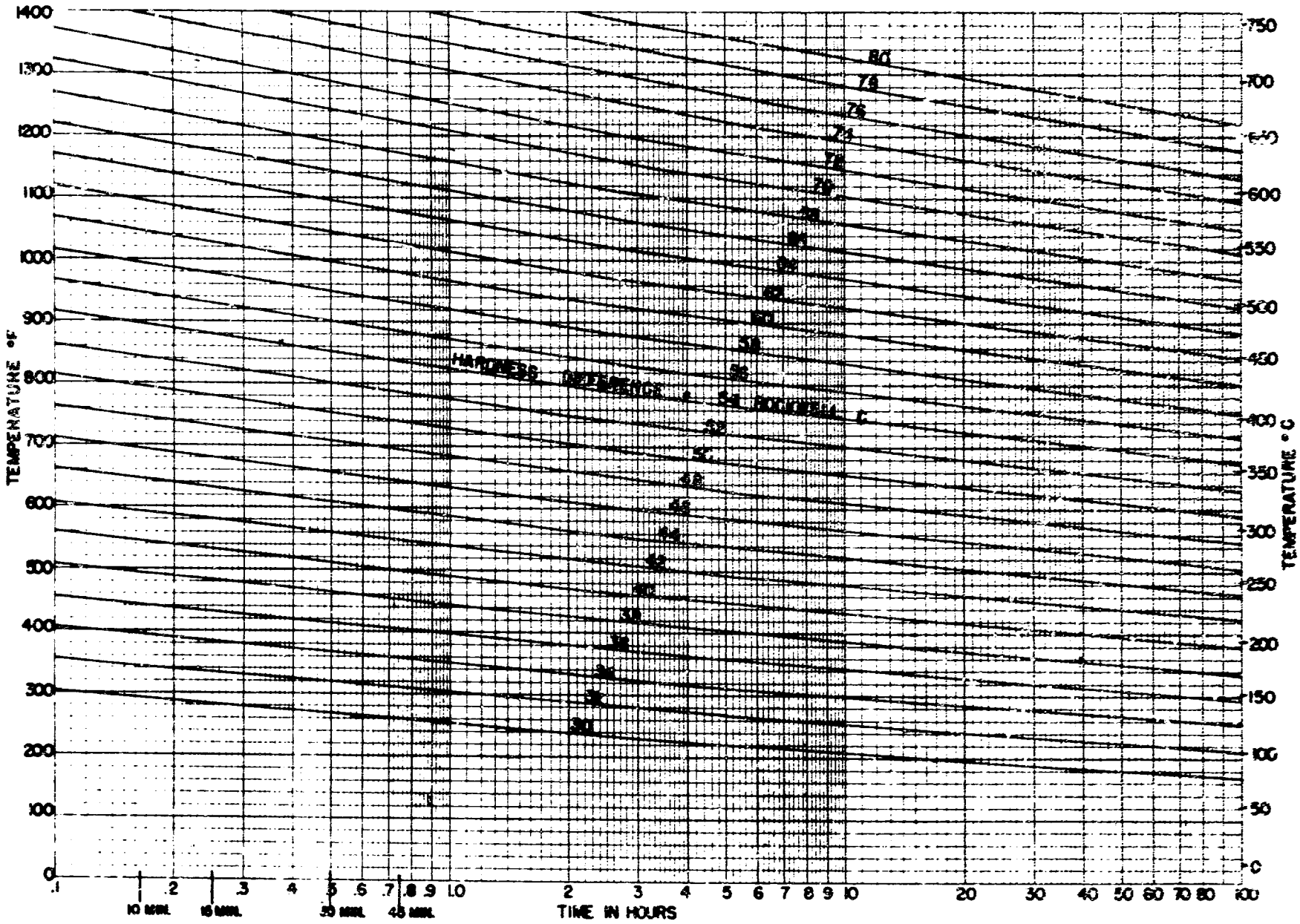
APPROXIMATE HARDNESS DIFFERENCES FOR STEELS

NOT CONTAINING LARGE PERCENTAGES OF

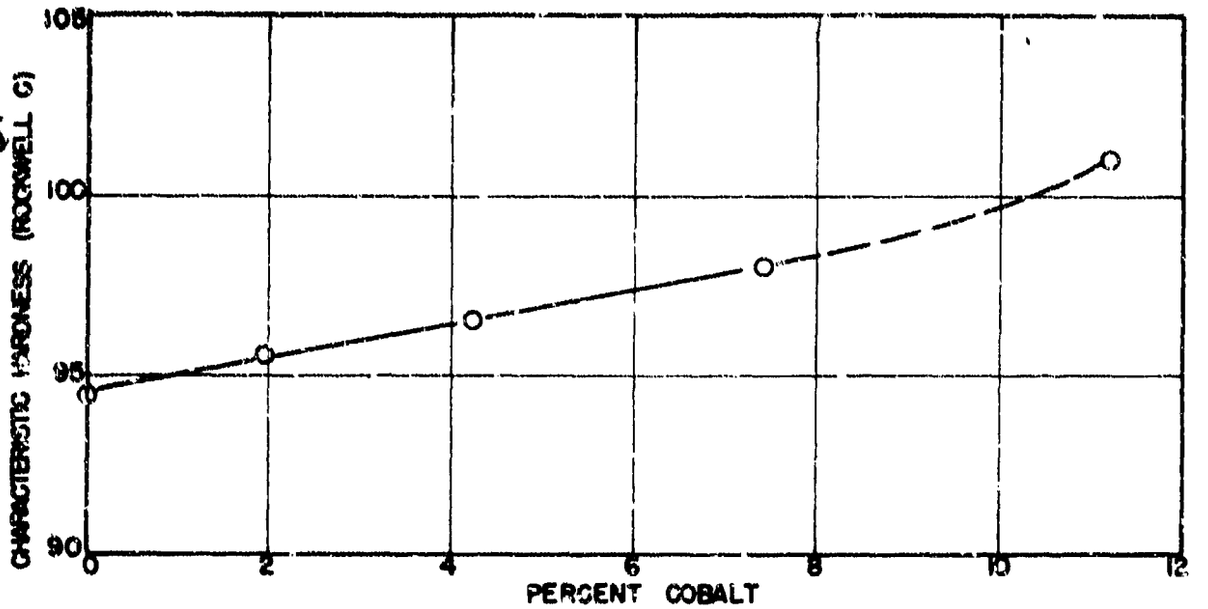
CARBIDE-FORMING ELEMENTS

ARE ALSO SHOWN.

c = 15 (TIME IN HOURS).


TEMPERATURE °F
TEMPERATURE °C
TIME IN HOURS
HARDNESS
ROCKWELL C
10 MIN.
15 MIN.
30 MIN.
45 MIN.

WTN.639-7427



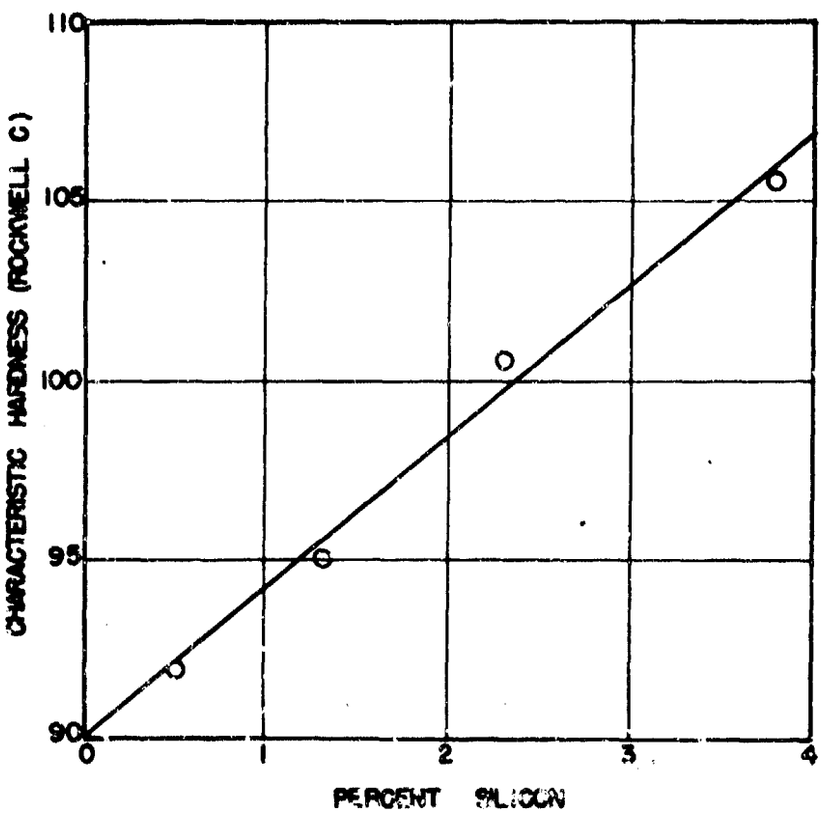


FIGURE 127

EFFECT OF COBALT AND SILICON UPON CHARACTERISTIC HARDNESS. (COBALT FROM DATA OF LORIA. SILICON FROM DATA OF BAIN[88])

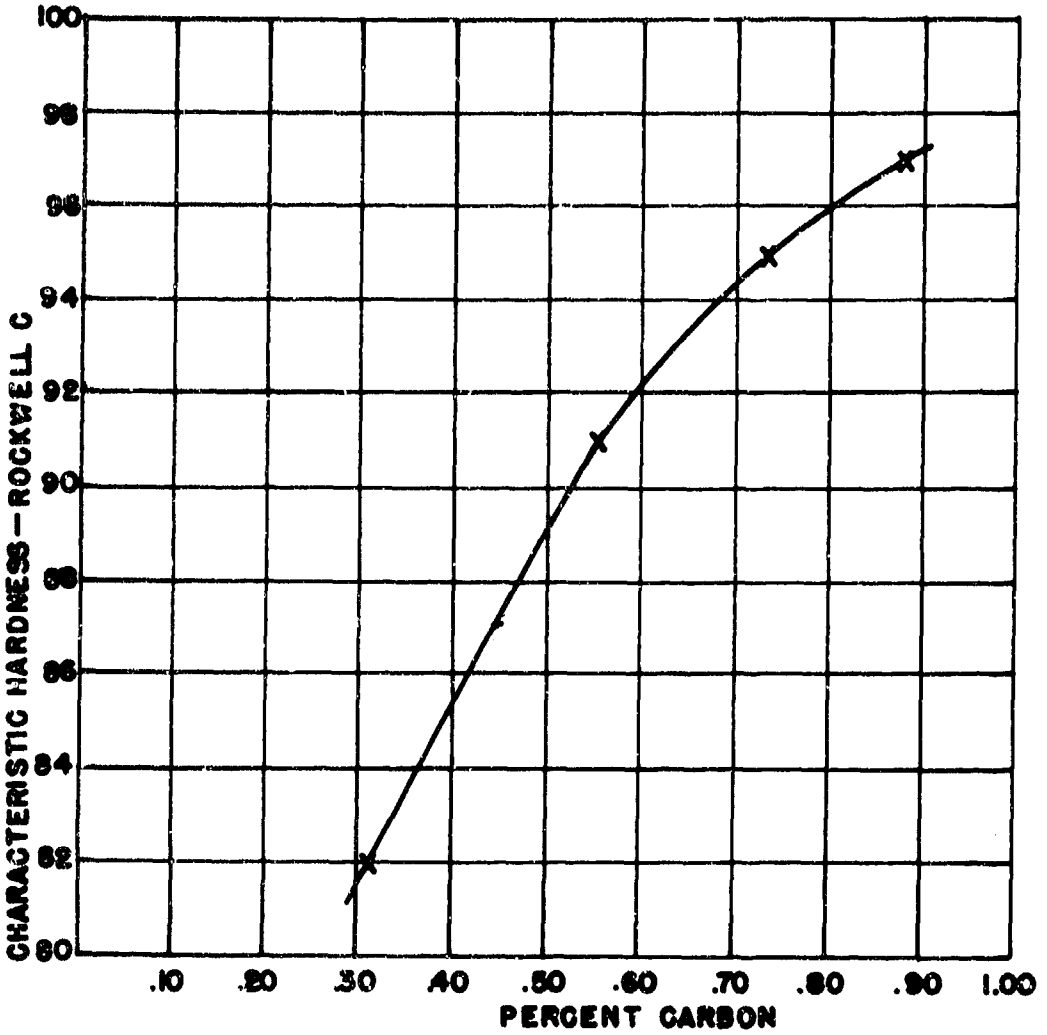


FIGURE 128

CHARACTERISTIC HARDNESS vs. CARBON CONTENT FOR IRON-CARBON ALLOYS

(FROM DATA OF HOLLOMON & JAFFE[122])

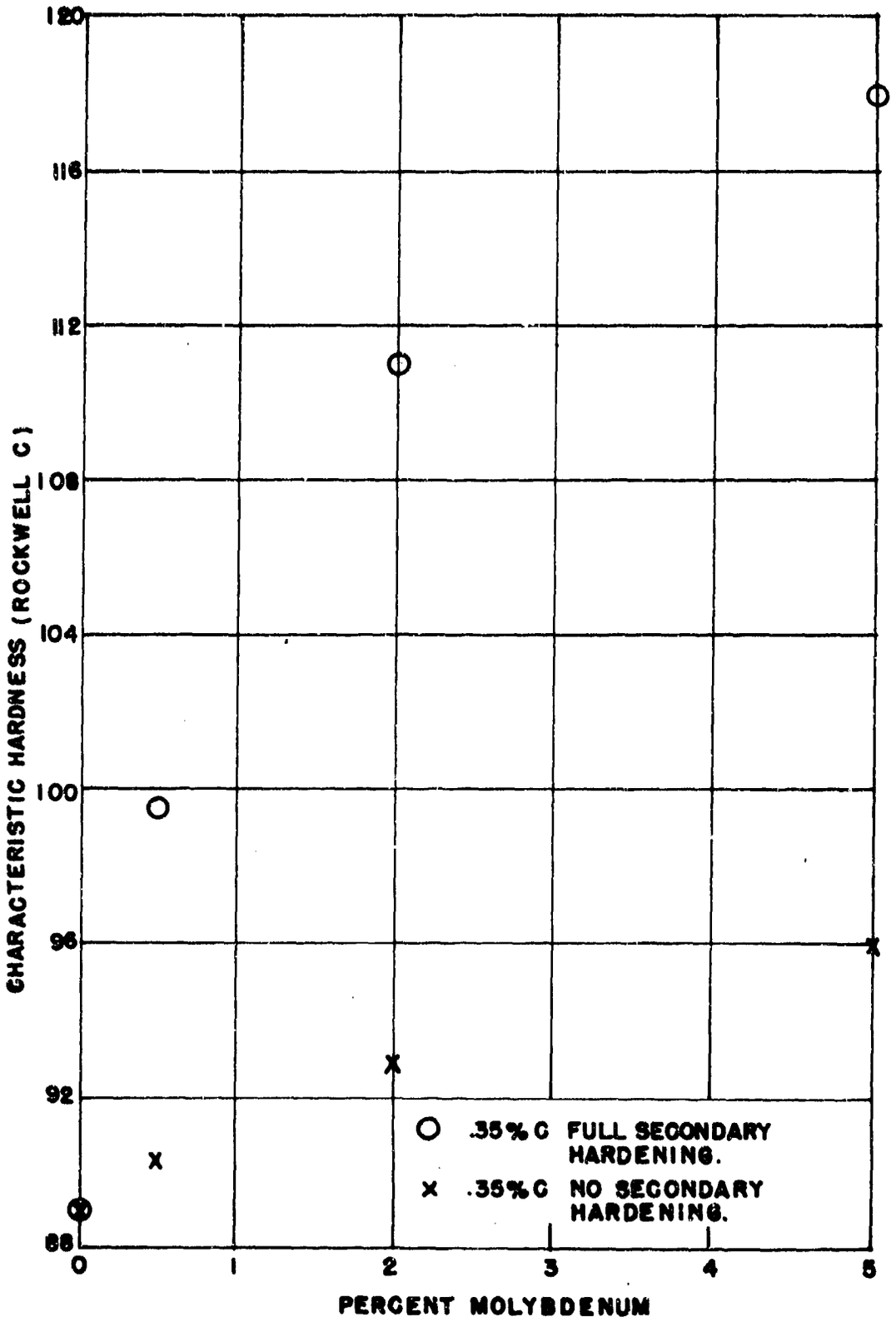


FIGURE 129

EFFECT OF MOLYBDENUM UPON CHARACTERISTIC HARDNESS (FROM DATA OF BAIN[250])

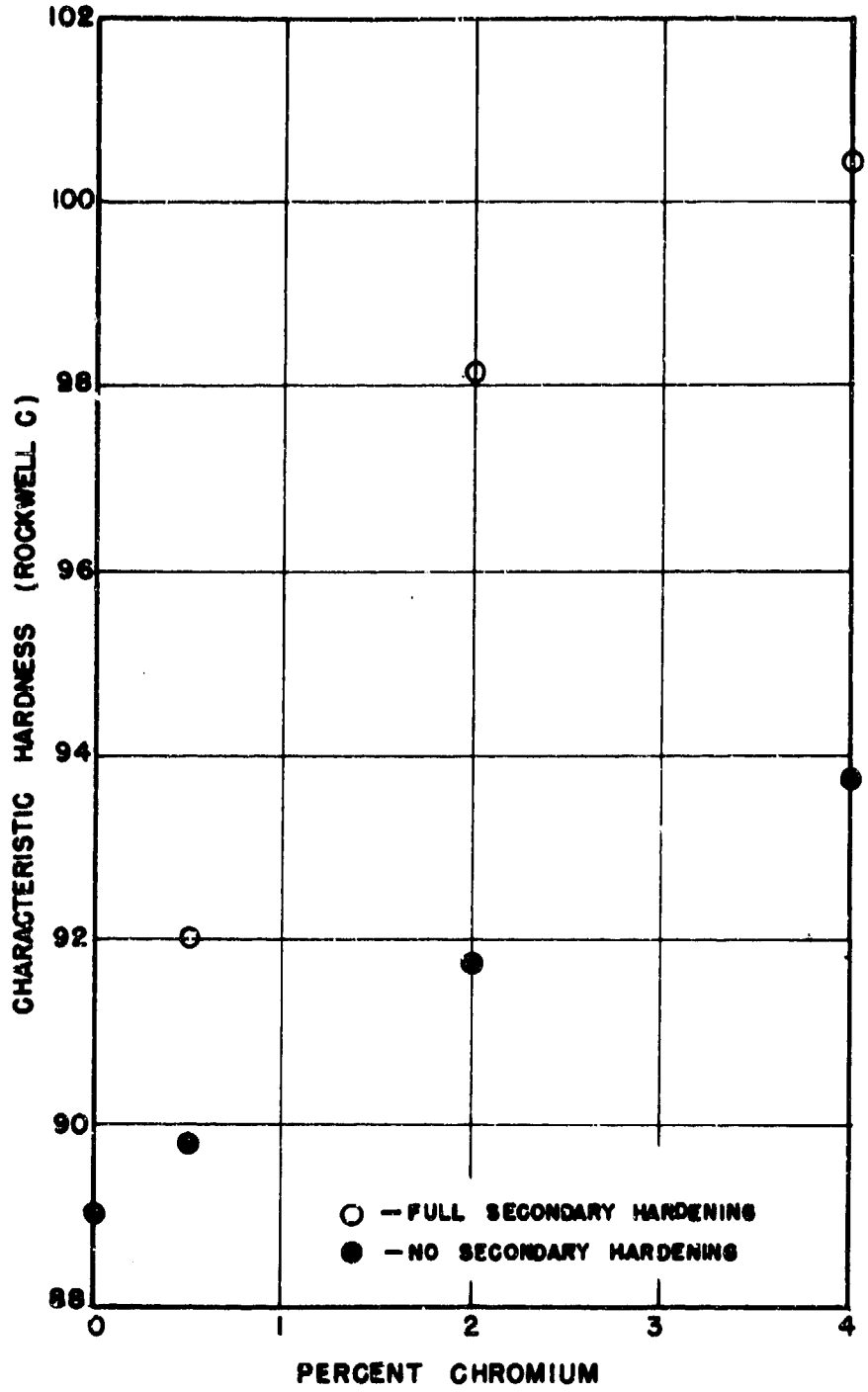


FIGURE 130

EFFECT OF CHROMIUM UPON CHARACTERISTIC HARDNESS (FROM DATA OF BAIN[290])

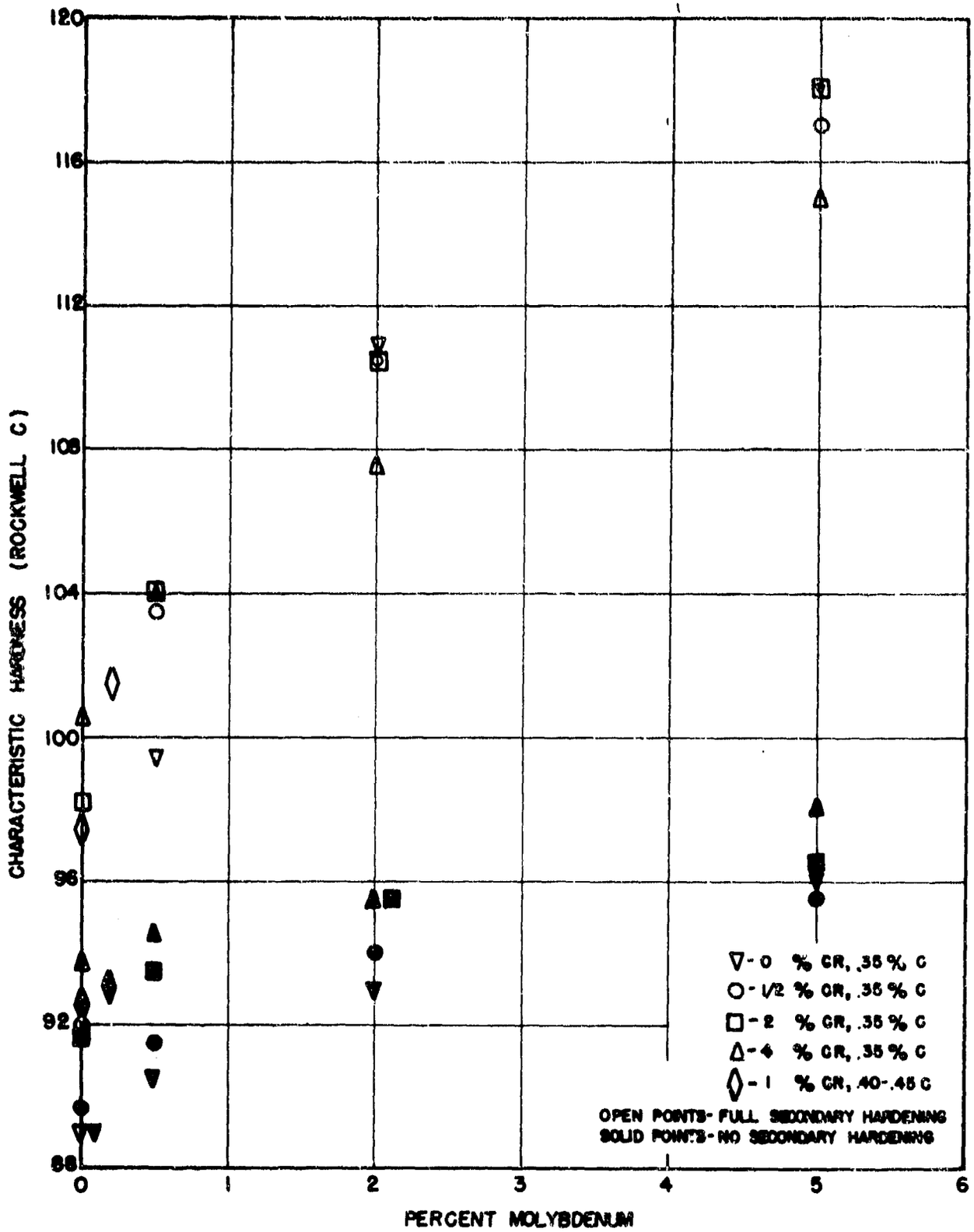


FIGURE 131

EFFECT OF MOLYBDENUM IN THE PRESENCE OF CHROMIUM UPON CHARACTERISTIC HARDNESS. (FROM DATA OF BAIN[299])

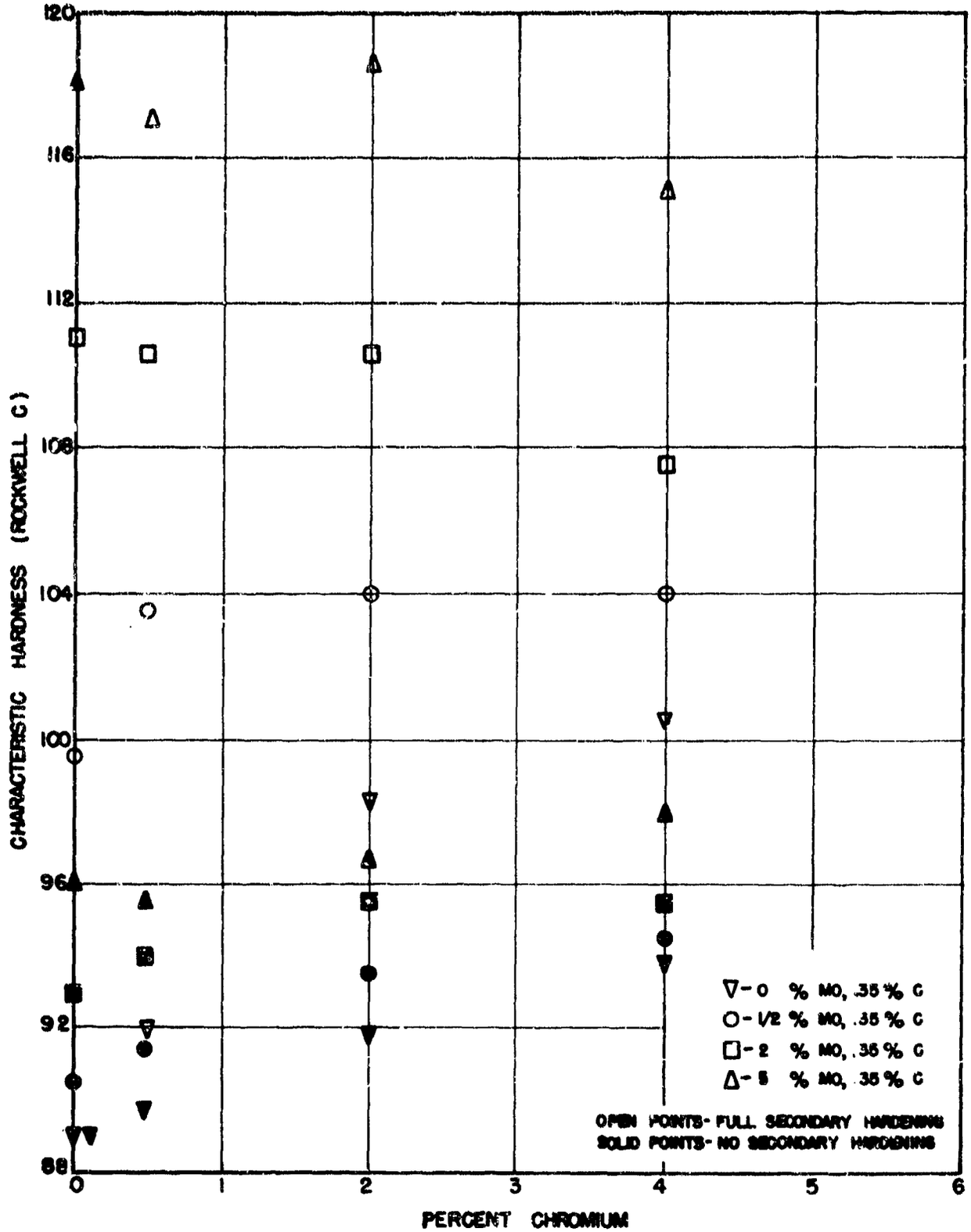


FIGURE 132

EFFECT OF CHROMIUM IN THE PRESENCE OF MOLYBDENUM UPON CHARACTERISTIC HARDNESS. (FROM DATA OF BAIN[296])

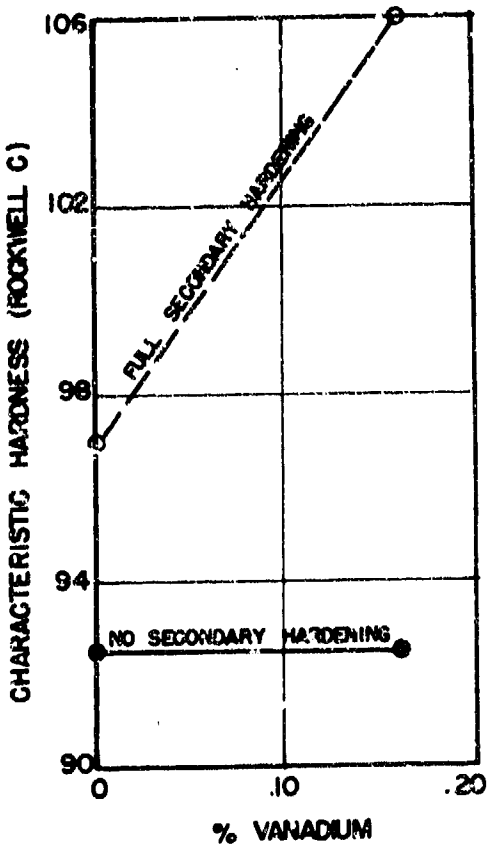


FIGURE 133

EFFECT OF VANADIUM

UPON CHARACTERISTIC HARDNESS

STEEL CONTAINING .40 - .45 %C, 1% CHROMIUM

(FROM DATA OF BAIN[299])

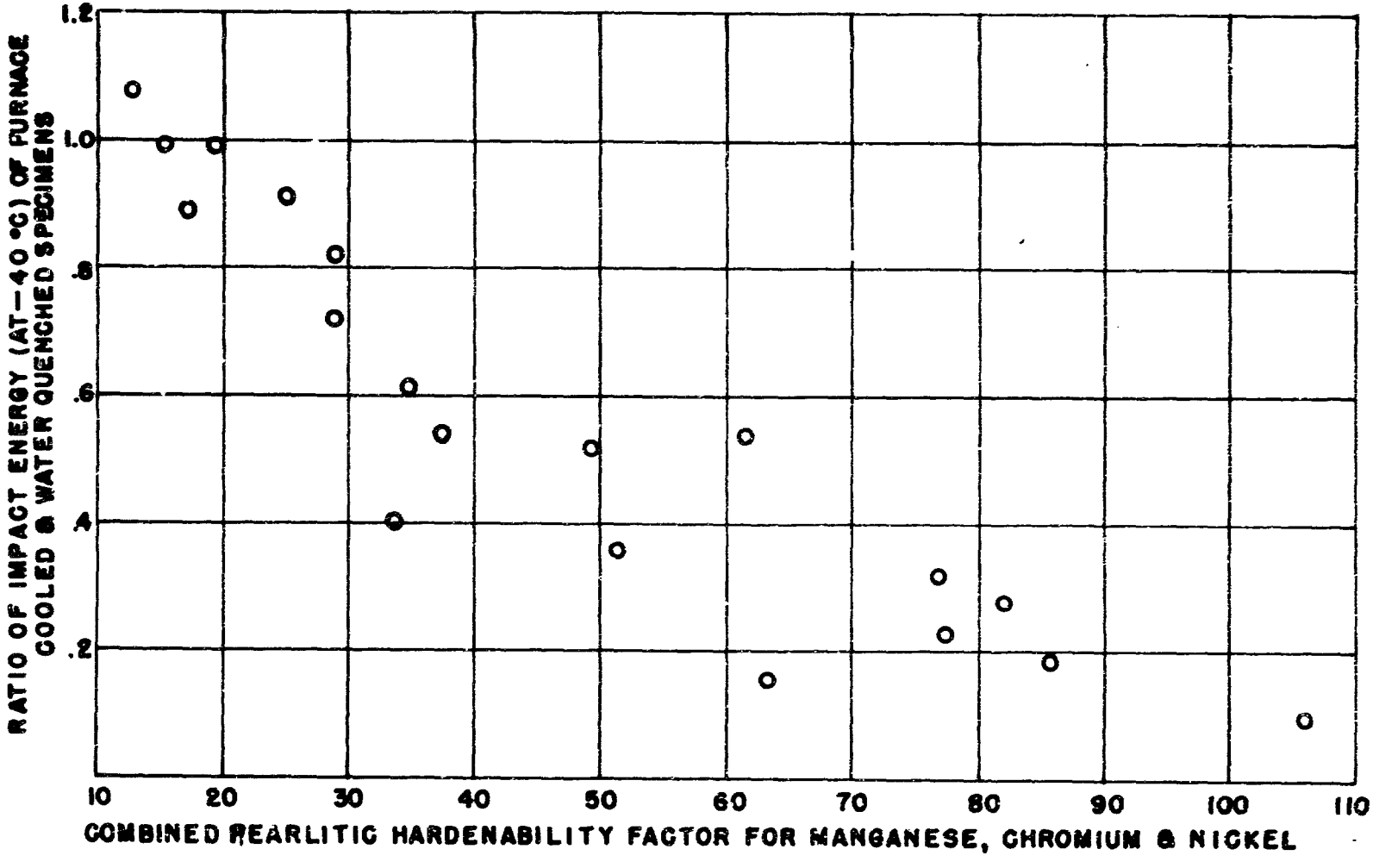


FIGURE 134

TEMPER BRITTLENESS SUSCEPTIBILITY OF STEELS OF VARIOUS HARDENABILITIES. (TESTED AT 26 ROCKWELL C, TEMPERED MARTENSITE).
(ALL CONTAINING .25 TO .40% Mo, 30%C, FURNACE COOLED AT 1/3°C PER MIN.) (AFTER HOLLOMON[189])

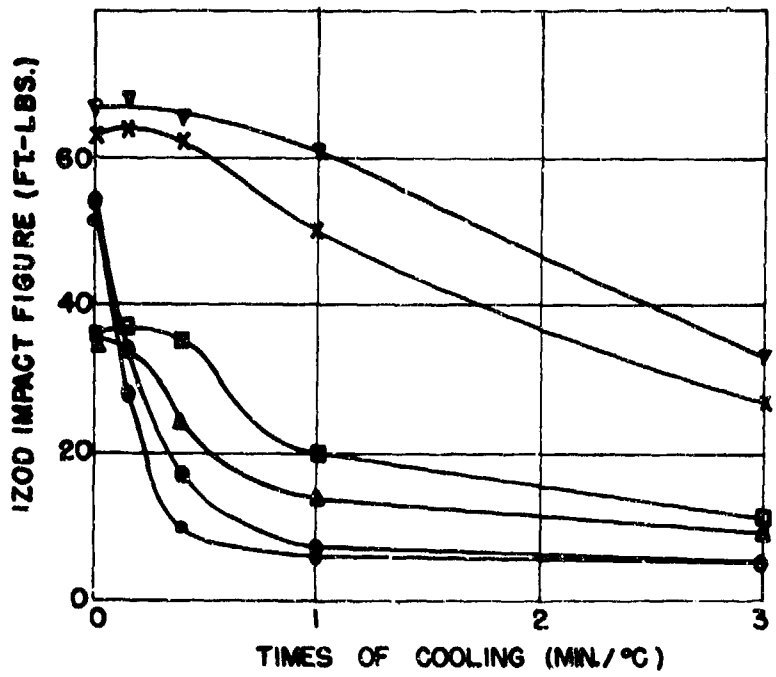


FIGURE 135

EFFECT OF RATE OF COOLING AFTER TEMPERING ON THE IMPACT FIGURE OF VARIOUS NICKEL-CHROMIUM STEELS. ROOM TEMPERATURE TESTS. (AFTER GREAVES & JONES[126])

SYMBOLS REPRESENT DIFFERENT NICKEL-CHROMIUM STEELS.

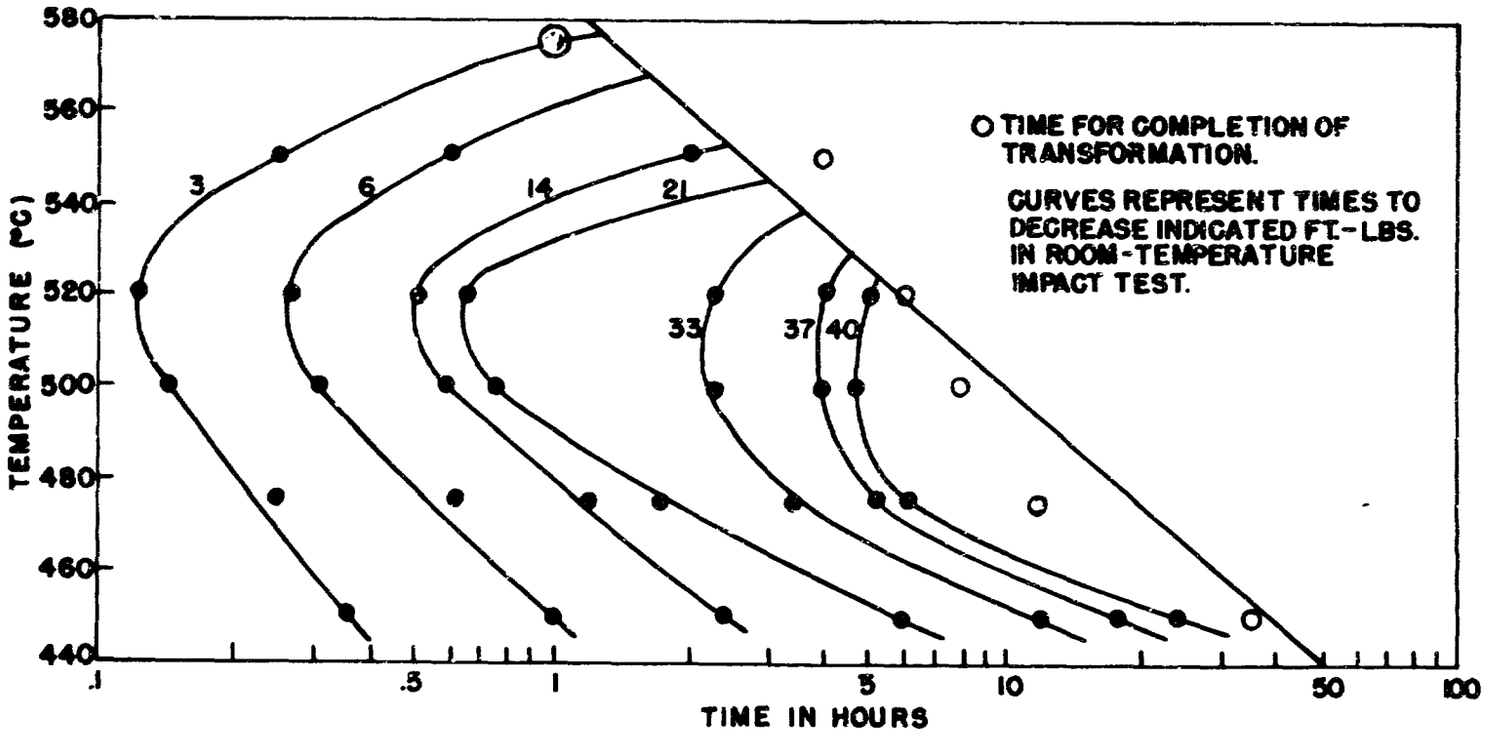


FIGURE 136

TIME-TEMPERATURE-TRANSFORMATION DIAGRAM FOR TEMPER BRITTLENESS PRECIPITATION. (AFTER HOLLOMON[129])

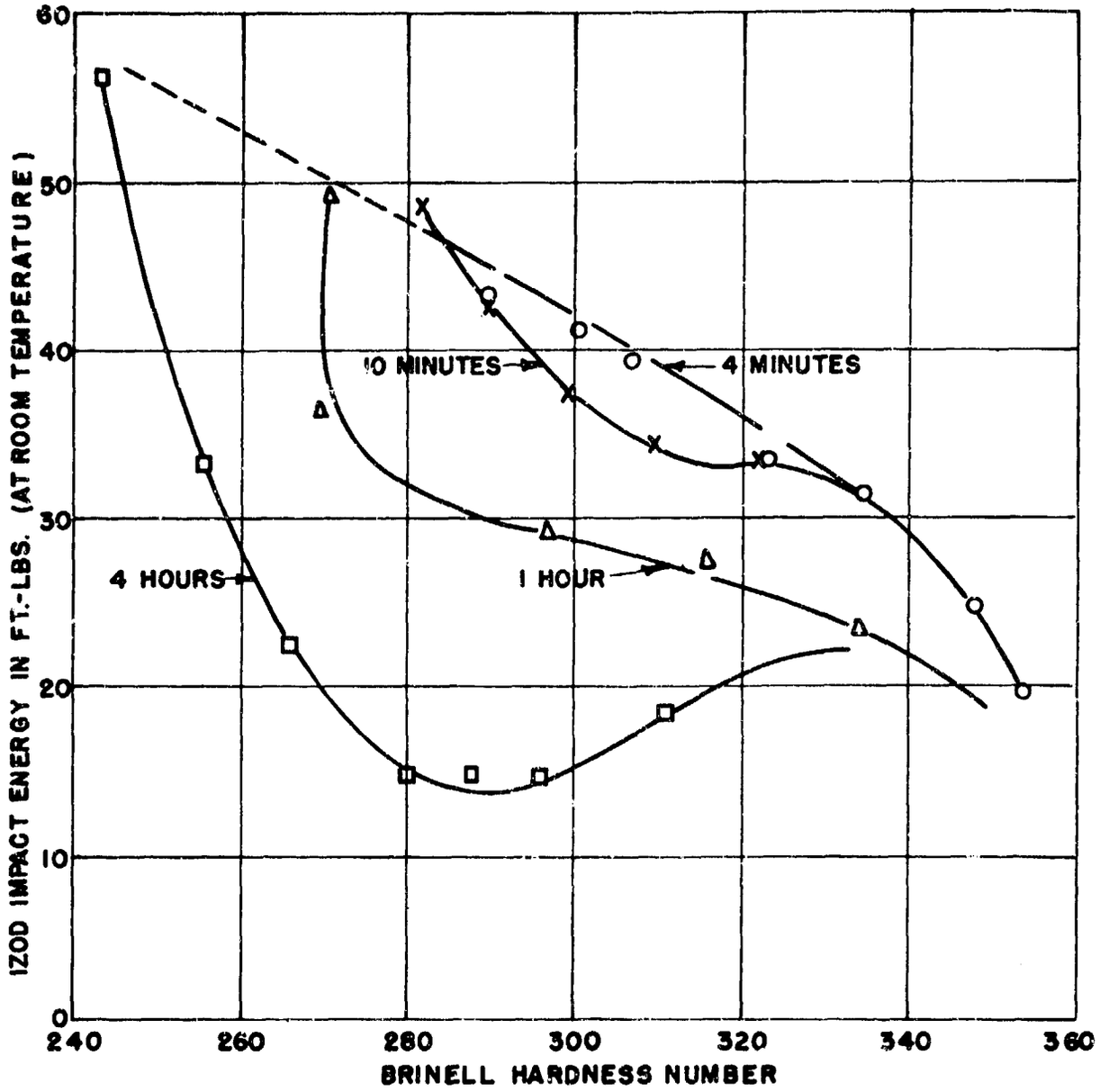


FIGURE 137

EFFECT OF TEMPERING AT VARIOUS TEMPERATURES FOR INDICATED TIMES FOLLOWED BY WATER QUENCHING (SUSCEPTIBLE NICKEL-CHROMIUM STEEL PRESUMABLY MARTENSITIC WHEN QUENCHED). (AFTER HOLLOMON[128]).

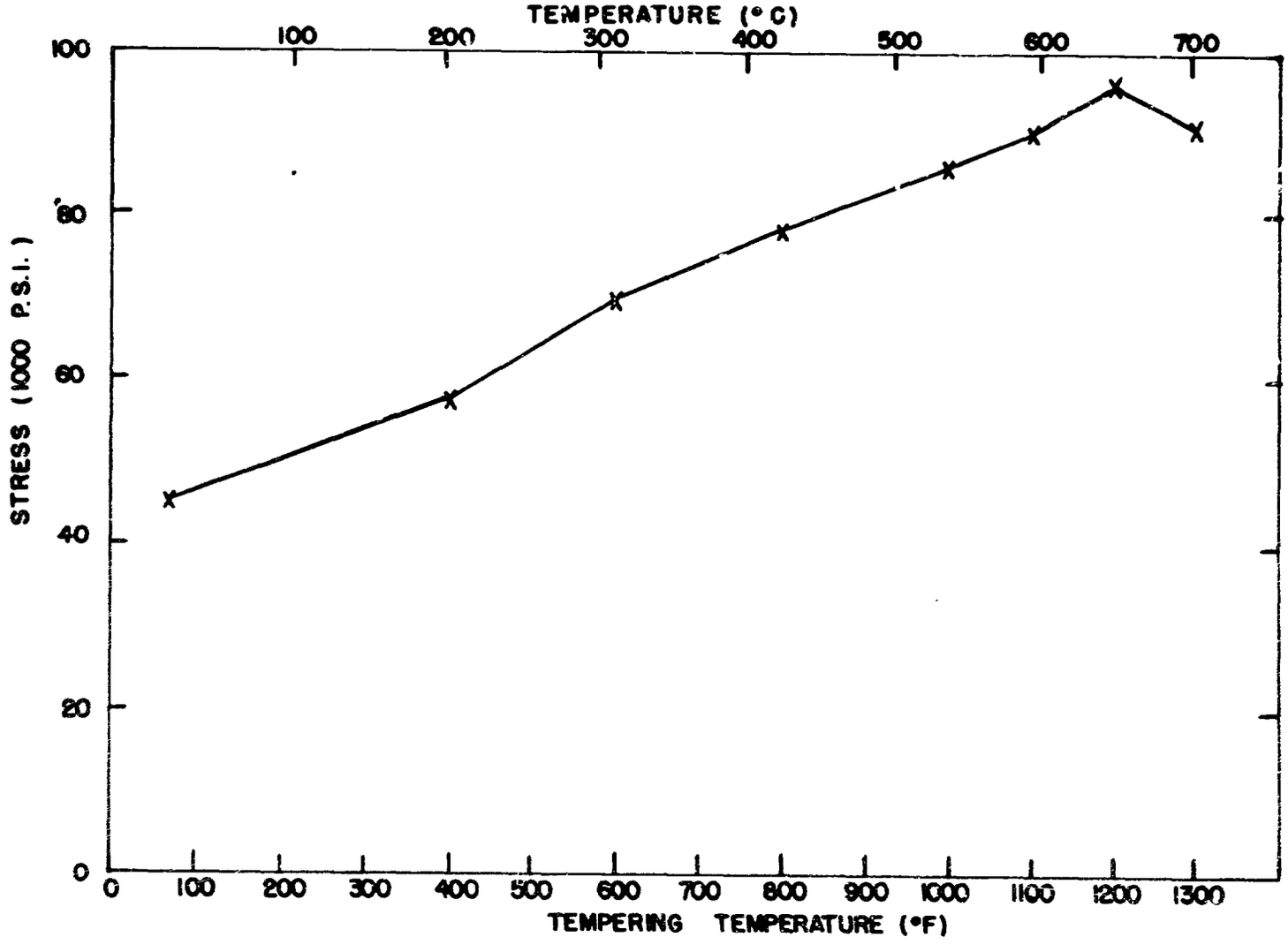


FIGURE 138

EFFECT OF TEMPERING TEMPERATURE ON THE .01% OFFSET YIELD STRENGTH OF A .20% CARBON STEEL. (AFTER PAPPAS & BENDER[303])

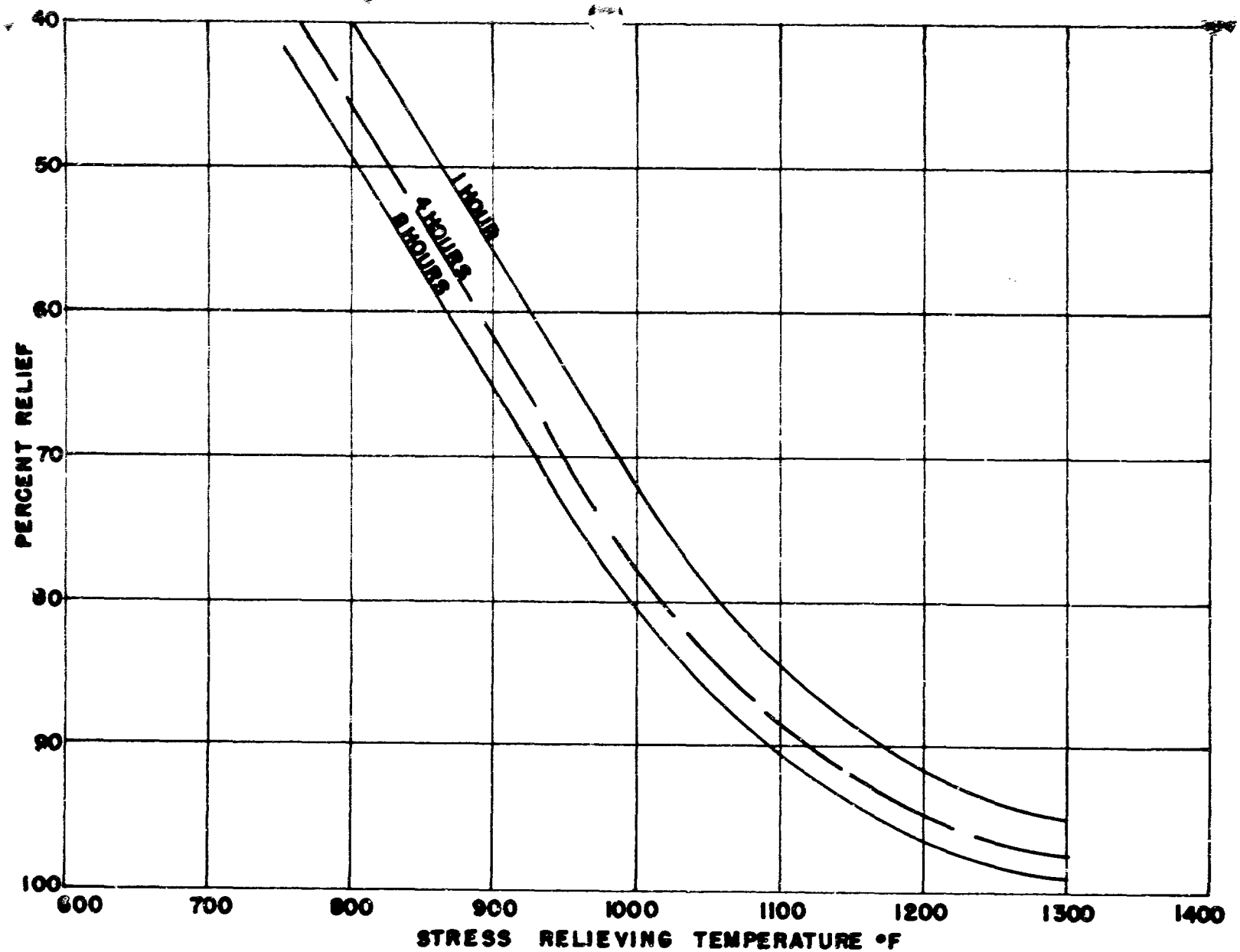


FIGURE 139

EFFECT OF TEMPERATURE & TIME ON THE RELIEF OF RESIDUAL STRESSES (AFTER JELM & HERRES[304])

Suggested Reading

E. C. Bain:
"Functions of the Alloying Elements in Steel", Am. Soc. Metals, Cleveland, 1939, pp. 228-295.

J. H. Hollomon and L. D. Jaffe:
"Time-Temperature Relations in Tempering Steels", Am. Inst. Mining Metal. Eng., Tech. Pub. No. 1831, 1945.

J. H. Hollomon:
"Temper-Brittleness", Am. Soc. Metals Preprint, 1945.

IX. DESIGN PROCEDURE

Using the knowledge of physical principles and the engineering principles and data presented in the earlier chapters, it is possible to develop a design procedure for obtaining the optimum mechanical properties in steel parts with minimum difficulty. As new and better information is developed, the exact heat treatments and compositions that will be designed may change, but the design procedure should not be altered radically. Even at the present time, the information available permits the design of heat-treatment and composition combinations on a rational basis. The first principle of such design is that optimum mechanical properties are obtained with steels having a tempered martensitic structure. The second principle is that all tempered martensitic steels have very nearly the same mechanical properties at a given hardness, provided their carbon contents are the same. A third, subsidiary principle, perhaps less generally applicable than the others, is that, in tempered martensitic steels of the same hardness, the ductility and toughness are greater the lower the carbon content. It is, therefore, generally only necessary for the mechanical designer to specify the shape and the hardness (or yield strength or tensile strength). The other mechanical properties are then approximately fixed. Thus, the metallurgical designer's problem is to obtain in the part a tempered martensitic structure of the desired hardness, with the lowest practical carbon content and without introducing metallurgical defects (such as temper brittleness, unfavorable internal stresses, etc.).

Before considering the design procedure further, it may be well to consider again the advisibility of obtaining a tempered martensitic structure. Such a structure does have optimum mechanical properties. However, if the steel is sufficiently soft, the separation between the flow and fracture curves for non-martensitic steels may be sufficiently great that their inferiority to tempered martensitic steels is not apparent under the conditions of service. Also, at high tempering temperatures and long tempering times, the structures of tempered bainite and even of tempered pearlite approach that of tempered martensite. If the strength required is low and high tempering temperatures can be used, it may not be necessary to obtain tempered martensite unless the service conditions are very severe. If, however, high strength is required and if tempering temperatures must be low tempered martensite will be necessary unless the service conditions are very mild.

It should also be noted that while tempered martensite may be needed it may not be needed throughout the section. Fracture generally begins in the surfaces of steel parts and it is the structure of the surface that is most important. Moreover, in parts hardened near the surface only, the stress introduced near the surface by quenching tend to be more compressive, and hence more favorable, than in parts hardened throughout. It must be pointed out, however, that as the hardenability decreases the percentage of the cross section of the part that will be fully martensitic decreases rapidly. Thus, slight variations in heat treating practice or in hardenability may radically change the amount of non-martensitic product formed and alter vastly the resulting properties. Generally, therefore, unless a very careful control of the hardenability and heat-treatment of

the steel is to be maintained, it is necessary to obtain a homogeneous structure. (In the same way, while small percentages of pearlite and considerably larger percentages of bainite may be tolerable in steels of moderate strength, tempered at moderate temperatures, and subjected to moderate service, it is very difficult to control satisfactorily the percentages of non-martensitic constituents.) Structures consisting of a surface layer of martensite on a non-martensitic core can be satisfactorily controlled through the use of surface-hardening procedures such as case-hardening (carburizing), flame-hardening, and induction hardening. While these procedures are satisfactory for many parts, they produce only a shallow layer of martensite.

If design is to be based upon a structure that is not essentially tempered martensite, the physical and metallurgical principles outlined above and certain elements of the design procedure described below may still be utilized. However, the procedure used will necessarily differ in certain respects from that for tempered martensite.

It was pointed out in the Introduction that design of a tempered-martensitic part involves design of shape and properties (hardness), design of heat treatment, and design of composition. It was also pointed out that each of these parts of the design problem is closely related to the other two, and that no one part can be considered complete until the other two parts are complete.

Design of shape and properties (hardness) is ordinarily the prime responsibility of the mechanical designer. However, cooperation between the metallurgical and the mechanical designers is needed in the design of shape and properties. For example, avoidance of quench-cracking must be borne in mind when the shape of the part is determined. In some cases, the hardenability required to obtain martensite in a heavy section may introduce such difficulties that it would be advantageous to make the part of two or more lighter sections. The maximum strength practical is generally limited by the low ductility and resistance to brittle failure of high-strength steels.

Design of heat-treatment and composition is the responsibility of the metallurgical designer. While several reviews of the factors involved in designing heat-treatment and composition have appeared(135, 305-307), little has been published concerning the sequence of design operations used to arrive at a satisfactory metallurgical design(278, 308-309). There are several logical sequences; the one suggested below seems the most feasible for designing a tempered martensitic part when the metallurgical designer has had little experience with that particular part. This sequence consists of the following steps:

<u>Step 1</u>: Determination of the general features of the quenching procedure to be used and the dimensions of the section that will cool most slowly.

<u>Step 2</u>: Estimation of the hardenability required to harden this section with this quenching procedure.

<u>Step 3</u>: Check of the estimate of required hardenability.

Step 4: Selection of the carbon level and type of steel to be used. Estimation of the level of alloying elements needed to provide the necessary hardenability.

Step 5: Estimation of the austenitizing procedure and the tempering procedure.

Step 6: Check of the estimated austenitizing procedure, hardenability, and perhaps tempering procedure for a series of steels similar to the one selected in Step 4.

Step 7: Manufacture of several parts from the steel that appears most suitable by the procedures that appear most suitable. Check for hardening, quench-cracking, temper brittleness, and properties.

Step 8: Readjustment of the design.

Step 9: Service tests of parts.

Step 10: Establishment of inspection and process control procedures.

Frequently, it is necessary, as the design proceeds, to stop and repeat previous steps before continuing. In fact, because the design of the heat treatment and the design of the composition are so interdependent, the choice of the final combination is actually a result of a series of successive approximations.

Step 1: Quenching

This part of the procedure consists of designing the quenching process and determining of the dimensions of the section that will cool most slowly. The quenching procedure cannot be definitely chosen until the composition is known and some sample parts have been heat-treated. It is desirable, however, to use the most severe quenching practice that does not introduce serious processing difficulties. Water (or other aqueous medium) is, therefore, generally the preliminary choice for a quenching medium. The quench may be interrupted, delayed, or stopped to reduce the tendency toward quench-cracking. These modifications generally need not be considered in the preliminary design of the quenching procedure.

For uniform quenching, it is usually desirable to move either the part or the medium. The method of applying the medium has a marked effect upon the severity of quench; the greater the speed of the medium, the more rapid and uniform will be the quench. Thus, if the equipment is available, spray or pressure quenching is advantageous. Extremely rapid quenches are possible with parts of simple shape, such as cylinders and plates. Past experience or a consideration of the stress concentrations present in the part may indicate that spray or pressure quenching is not feasible, but these practices often permit the quenching of recesses otherwise difficult to cool satisfactorily.

After a tentative quenching procedure has been chosen the position in the part that will cool most slowly can be estimated. The dimensions of the section that will cool most slowly are fixed primarily by the dimensions of the finished part and the allowance for machining or grinding after heat-treatment. The most slowly cooled section will usually be the thickest section, but cooling from the edges or ends must be considered. Since decreasing the thickness of the most slowly cooled section decreases the hardenability that is necessary, it is advantageous to perform as much as possible of the rough-machining of this section before quenching. However, threads and sharp fillets may well be machined after heat-treatment to avoid cracking during the treatment. Of course, the hardness before and after the heat-treatment will affect the practicality of machining in each condition.

Step 2: Estimation of Required Hardenability

As the hardenability of steel is generally expressed in terms of ideal round size or Jominy distance, the Jominy distance or ideal round size having cooling conditions equivalent to those of the most slowly cooled sections should be estimated. The charts and information for this estimation are included in Chapter V. As indicated in that chapter, the conversions recommended are approximate only.

Both the pearlitic and the bainitic hardenability required should be estimated. If direct or interrupted quenching is to be used, the quenching severities, and hence the required hardenability, may usually be considered the same for both pearlite and bainite ranges. If delayed or stopped quenching is planned, then the estimate of pearlitic hardenability should be based upon the severity of the medium to be used for quenching through the pearlite range, and the estimate of the bainitic hardenability upon the severity of the medium to be used for quenching through the bainite range.

Step 3: Check of Estimate of Required Hardenability

For precise results, it is desirable to check the estimate of the required hardenability*. The suggested method is to make a sample of the part from a steel that will harden only partially, and quench the part using the procedure under study. The quenched part is then sectioned in the as-quenched condition at the most slowly cooled position, and hardness test or microscopic examination made on this cross section. The results can be then compared with the results of a standard hardenability test performed in the same heat of steel. The method is most sensitive if the sample part is made of a steel that will contain ten to fifty per cent martensite at the most slowly cooled position. (The steel to be used for the sample can be selected by the methods outlined in Step 4.)

The sample part, machined to the dimensions in which the production

*It is possible to omit either Step 2 or Step 3 from the sequence, but not both.

parts are to be heat-treated, should be austenitized and then direct-quenched*. To obtain the same severity of quench as is to be used for production parts, the same quenching tanks, pumps, and fixtures should be employed. After quenching, the part should be sectioned to expose the area where the slowest cooling was expected. If there is any doubt as to the location of the most slowly cooled area, sufficient sections should be taken to expose all the possibilities. It is necessary to section the part in the as-quenched condition, since after tempering it is difficult to determine the exact degree of hardening by either hardness or microscopic tests (at least when bainite is present). The sectioning can perhaps best be carried out with an abrasive cutoff wheel, using a large quantity of coolant. It can also be done by nicking the piece, breaking it, and wet-grinding the fractured surface. Another method is to flame-cut the piece and then grind off enough material to remove the heated zone. The steel to be examined must, in any case, be kept cold throughout the whole process.

After proper preparation the surface can be tested for hardness or examined microscopically. Hardness tests are, of course, the simpler to perform. To facilitate comparison with Jominy readings, the Rockwell C scale can be used advantageously. Sufficient readings should be taken to insure an accurate determination of the hardness of the softest portion of the piece. Similarly, when the microscope is used, the per cent of martensite at the portion containing the smallest per cent should be determined.

The hardness or structure found in the part must then be compared with that of the hardenability test bar. The Jominy test or, for steels too shallow-hardened for accurate Jominy testing, one of the other hardenability tests standardized by the Society of Automotive Engineers and the American Society for Testing Materials, is recommended (Chapter VI). The steel used for the hardenability test should be of the same heat as that used in the sample part. The steel of the test bar should be given the same preliminary heat-treatment (normalizing, annealing, etc.) as the sample part and austenitized at the same temperature for the same time**. The test bar should then be quenched in the standard manner, and the hardness or microscopic survey made. The method (microscopic or hardness) should be the same on the test bar as on the sample part. By comparison, the position on the test bar (the hardenability) equivalent to that at the least-hardened portion of the sample part is located.

---

*It is not possible to reproduce in a directly-quenched hardenability test bar the conditions and structures obtained by delayed, stopped, or interrupted quenches. Therefore, Step 3 is not directly applicable when such quenches are used. Modifications necessary to cover more complex quenches can be determined by making pilot parts of steels modified from the direct-hardening composition in accordance with the principles of Chapter VII and testing them as outlined in Step 6.

**This means that the twenty-minute austenitizing time prescribed in the standard Jominy procedure should be disregarded. Moreover, if long austenitizing times are used, special precautions may be needed to avoid the effects of surface decarburization or carburization. Thus, Jominy bars may have to be ground .050 inch deep or more, instead of the standard .015 inch.

The air-cooled end of the Jominy test bar is approximately equivalent to the center of a 3-1/2-inch plate or a 5-1/2-inch round quenched in still water, and to a 2-inch plate or a 3-1/2-inch round in still oil. When larger sections are of interest, the standard Jominy test cannot be used. Standard hardenability tests for deeper hardening steels have not yet been adopted, but as stated in Chapter VI, some have been recently suggested. When the section is thick (say over 3" plate or 5" round), it may be advisable to omit Steps 3 and 6 and try a series of steels (selected by Step 4) in Step 7.

Step 4: Selection of Composition

Since carbon decreases the toughness of the steel and has such a tremendous effect upon the tendency toward quench-cracking and the retention of austenite, the carbon content should be kept as low as possible. The exact limit of carbon is determined by the amount necessary to produce the required strength with the necessary tempering treatment. Frequently, a minimum tempering temperature is necessary to permit stress-relieving after welding or cold fabrication or to avoid temper brittleness*. Furthermore, it is often considered necessary to temper above about 900° F. (475° C.) to relieve the quenching stresses and to straighten the stress-strain curve. After the limitations to the tempering temperatures have been established, the minimum carbon content should be chosen that will give the desired strength with the minimum possible tempering temperature. The carbon content necessary to produce the desired strength with this temperature may be estimated by the methods of Chapter VIII. Fortunately, the alloy additions necessary for hardenability have, compared to the carbon content, relatively little effect upon the hardness of the tempered steel, unless the additions are large.

After the carbon content has been selected, alloy contents that will provide the required hardenabilities must be chosen. Methods for estimating the pearlitic and bainitic hardenabilities provided by a steel composition were discussed in Chapter VI. There are, of course, an infinite number of combinations of alloying elements that will give the hardenabilities necessary. Other considerations limit the choice. Limitations or requirements of refining, casting, or forming practices often affect the level of some of the alloying elements. For example, it may not be practical to make high-chromium steels in acid-lined furnaces. It may not be practical to roll high-silicon steels. Some elements may have specific effects upon the steel. Thus, the presence of sulfur and phosphorus is generally believed to have detrimental effects upon the toughness, and these elements are, therefore, ordinarily kept as low as possible. Silicon contents over .50 per cent have also been reported to lower the toughness, at least in cast steels. Molybdenum (up to at least .40 per cent) decreases the speed with which temper-brittleness appears. Very high percentage of

*The susceptibility to temper brittleness cannot be estimated until the analysis is chosen, but some idea, at least for steels containing .30-.40 per cent carbon, can be obtained from the hardenability required (Figure 134).

carbide-forming elements are likely to result in formation of carbides that will dissolve in a reasonable time only at very high austenitizing temperatures, if at all. Price considerations naturally enter. Frequently, it is very desirable to minimize the tendency toward quench-cracking. In such cases the method of choosing the alloying elements described in Chapter VII is likely to prove useful. Special addition agents (boron, zirconium, titanium, vanadium) are not considered in that method. They may be used if their effects are understood and are believed to be controllable. A summary of the relative effects of the different alloying elements upon the various factors influencing design is given as Table VIII.

The choice of the so-called "deoxidation practice" depends on the method of steel-making. However, if large percentages of carbide-forming elements are to be used, aluminum additions are usually necessary to prevent excessive grain growth at the high temperatures that must be used for austenitizing.

If the part is to be cast, there is generally little difficulty in obtaining steels of the desired composition. Furthermore, if a large tonnage of parts is to be made from rolled or forged stock, it is possible to order special heats made to the desired chemistry and hardenability. For small tonnages of parts, it may be necessary to use readily available commercial steels. These steels generally have not been designed on metallurgical principles but rather have been chosen on the basis of production facility and experience. The general design principles, may, however, be applied to the choice of a commercial steel. In Table V the pearlitic and bainitic hardenabilities (ideal round) for the average compositions of SAE and NE steels are listed. That composition which is best suited can be chosen. Except at high carbon contents, no steels suitable for heavy sections are included. It should be remembered that the hardenability of commercial steels varies widely within the commercial composition ranges, and this variation should be taken into account in choosing the steel.

Step 5: Estimation of Austenitizing and Tempering Procedures

After the final estimate of composition has been made, the tempering-treatment may be estimated anew and, if it is possible, the carbon content should be reduced and the analysis readjusted. The proper balance between tempering-treatment and composition should be approached by a series of successive approximations, always maintaining the necessary pearlitic and bainitic hardenabilities.

An estimate of the susceptibility of the steel to temper-brittleness should be made from the data of Chapter VIII (or on past experience), and the compositions adjusted, if necessary, to minimize temper-brittleness. For susceptible steels it is desirable to quench from the tempering temperature and to temper above 1100° F. (600° C.). Adjustments may be necessary to attain the desired strength with this tempering temperature. A reduction of the tempering time may be in order.

An estimate should be made as to whether the tempering temperature will exceed the lower boundary of the equilibrium A_1 range. As indicated in Chapter I, manganese and perhaps nickel, for example, lower the

A_1 range, while molybdenum and probably chromium raise it. With steels moderately high in manganese and nickel, lower critical temperatures below 1200° F. (650° C.) are not uncommon. Adjustments may have to be made to avoid tempering in or above the A_1 range, and the resulting likelihood of austenite formation on tempering.

Because of the inferior fracture strengths produced by tempering treatments equivalent to 1 hour at 450-650° F. (250°-350° C.), such treatments are also to be avoided.

The austenitizing temperature necessary to obtain solution of the carbides should be estimated on the basis of the chosen analysis. The proper austenitizing temperature is determined not only by the equilibrium diagram, but also by the time required for solution and homogenization. If a short austenitizing time is selected, the temperature must be higher than if a long time were to be used; however, slight increases in temperature will compensate for large decreases in time. The austenitizing time is usually a matter of operating convenience for the production parts; from a theoretical point of view, it need be only long enough to insure reproducibility in the treatment of successive pieces. The time the steel is at temperature, rather than the total time in the furnace, is the important factor. In general, the higher the alloy content, particularly the content of carbide-forming elements the higher the austenitizing temperature will have to be.

It may also be advisable at this point to make a rough estimate of the martensite temperature range. An estimate can be made from composition on the basis of Table I or Table VI. This will give an indication of the temperature to which the steel must be cooled to transform the austenite to martensite, and of the likelihood of austenite retention if the steel is cooled only to some higher temperature.

Step 6: Check of the Austenitizing and Tempering Procedures and Hardenability

It is usually desirable to check the estimate of the composition that will give the required hardenability and of the austenitizing and tempering procedures to be used with it. It is advisable to test a series of steels, similar to that tentatively selected, varying in hardenability by more than the probable errors in the hardenability estimates — say varying by at least 25 per cent (in terms of ideal round). This variation can perhaps best be obtained by preparing heats differing systematically in one or two elements.

Before determining the hardenability of the selected steels, it is necessary to make a final selection of the austenitizing cycle, which means choosing the preliminary heat-treatment, the austenitizing time, and the austenitizing temperature. The preliminary heat-treatment is generally given to facilitate machining and, in some cases, cold forming or straightening. This treatment can usually be definitely established at this point.

The selection of the austenitizing temperature can be checked by preliminary tests. One method consists of quenching small pieces of the

steels under consideration from a series of temperatures, using the selected preliminary treatments and austenitizing times, and examining them microscopically for undissolved carbides or ferrite. The austenitizing temperature can then be taken as that at which the carbides and ferrite disappear plus perhaps 50° F. (30° C.) for safety. Austenitic grain size can be checked at the same time, by either fracture or microscopic methods. Fine-grained steels are generally desired because they are considered to have greater toughness. With proper aluminum deoxidation, steels will ordinarily remain fine-grained at the temperature needed for complete austenitizing. Instead of preliminary tests on small samples, several hardenability test bars can be used for each steel and austenitized at a series of temperatures (after the chosen preliminary treatment and for the chosen austenitizing time). After quenching, the bars can either be examined for carbide solution or for hardenability. In the latter case, hardenability is plotted as a function of temperature and the temperature at which hardenability ceases to increase sharply is, after adding perhaps 50° F. (30° C.) for safety taken as the austenitizing temperature. If this method is used and pearlitic hardenability limits the hardening, care should be taken to exclude the effect of possible changes in austenitic grain size upon the pearlitic hardenability. This effect can be calculated by the use of Table IV.

To check the estimate of the composition that will give the required hardenability, a hardenability test should be made on each of the selected steels. The type of test used in Step 3 should be used here also. If Step 3 was omitted, any hardenability test whose correlation with the shape and quenching procedure is well known may be used. (However, if Step 3 was omitted because of thickness of section or for other reasons, it may be well to also omit hardenability testing from the present step and test the series of steels by making sample parts from all of them in Step 7.)

The steels should be given the selected preliminary treatments and hardenability test bars then prepared, given the selected austenitizing treatment, and quenched in the manner standard for the test. The extent of hardening may then be determined by one of the methods discussed in Chapter VI. The metallographic method is most reliable and is the only one that indicates whether pearlite or bainite formation limits the hardenability. On the basis of the hardenability results, the composition that appears most suitable can be selected for future work, or several compositions can be selected for further consideration.

The behavior of the selected steel or steels during tempering can well be checked at this point. The hardness can readily be determined at several tempering temperatures. The effect of time can then be evaluated by the use of Figures 125 or 126. Through use of the principles given in Chapter VIII, considerable information as to the temperability can be obtained with little experimental work.

If it is expected that temper-brittleness may be encountered in the part, appropriate checks may be made. Such checks are ordinarily carried out by making comparative notched-bar tests upon specimens that would be expected to differ in the amount of temper-brittleness precipitate present. The notched-bar tests are of value only if made at a temperature within or

below that at which the most embrittled specimens tend to fracture brittlely (Chapter IV).

If the tentative tempering temperature is not much above the nose of the temper brittleness C-curve (which is about 975° F. (525° C.)), the notched-bar properties obtained with the selected treatment may be compared with those obtained on specimens tempered to the same hardness, using a higher temperature and a shorter time. (The latter specimens should be small and should be quenched from the tempering temperature.) If a difference is found at the proper testing temperature, the selected tempering treatment embrittles the steel. If the above procedure is not practical or if the selected tempering temperature is considerably above the nose of the C-curve, all that is worth doing in this step is to determine whether the steel is susceptible to temper-brittleness. This can be done by tempering small specimens above 1100° F. (600° C.) and quenching from the tempering temperature. The notched-bar results on these specimens may be compared with those on similar specimens given a subsequent treatment that would cause embrittlement in a susceptible steel. Cooling from the tempering temperature at a slow rate, such as 1/3° F. (1/2° C.) per minute, or holding at a temperature of about 850° F. (450° C.) for a moderate time, such as 5 hours, is suitable.

If it appears that temper-brittleness is likely to be troublesome, suitable adjustments in tempering treatment and in composition should be made. Raising the molybdenum content decreases the rate at which embrittlement occurs. Raising the tempering temperature, up to 1100° F. (600° C.), decreases the likelihood and extent of temper-brittleness. This may be done by shortening the tempering time, or the resistance to tempering may have to be increased. Tempering in lead or salt or by induction offers possibilities of using very short times. Increase in resistance to tempering may be accomplished by raising the carbon content or by the introduction of, particularly, carbide-forming alloying elements (Table VII). The increase in carbon content also decreases the amount of alloy necessary for hardenability and so permits a decrease in susceptibility to temper-brittleness. The improvement with regard to temper-brittleness must be balanced against the disadvantages of higher carbon or alloy content. If temper-brittleness is to be minimized at a tempering temperature above the nose of the embrittlement C-curve (approximately 975° F., 525° C.), cooling from the tempering temperature should be as rapid as possible.

Step 7: Check of the Selected Composition and Heat Treatment Combination

One or more sample parts should be made from the selected steel or steels, after the selected preliminary treatments. These parts should be processed through the hardening operation in the way proposed for the production parts. The sample parts should be checked for quench-cracking and undue distortion. Magnetic powder tests or other tests may be necessary to discover the quench-cracks.

The parts should then be examined to see if they have been fully hardened. There are several ways of doing this. One method is to section the piece in the quenched condition and determine the hardness or, preferably, the microstructure in the most slowly cooled portion. Another method

is to temper the part to a hardness of from 260 to 350 Brinell (preferably 260 to 300) and then notch the part and fracture it through the critical area. A "fibrous" fracture is indicative of essentially complete hardening. "Crystallinity" in the fracture may show incomplete hardening. Also, after tempering to this hardness, results of tensile and notched-bar tests taken from the critical areas and from portions of the part known to be fully hardened can be compared. The notched-bar energies may also be compared with the values shown in Figures 75 to 77. Incomplete hardening (and other metallurgical defects*) are indicated by "crystallinity" in the fracture tests, low yield-tensile ratios in the tensile tests, and low impact energies and "crystallinity" in the notched-bar tests. These tests are more sensitive when made at subatmospheric temperatures. However room-temperature tests are usually sufficient at the indicated hardness level. If poor results are obtained, microscopic examination of the part in the untempered condition is the surest way to determine whether they are due to incomplete hardening or to some other defect. If the parts are found to be incompletely hardened, adjustments must be made as indicated below (Step 8).

If the parts are found to have been fully hardened, the choice of the tempering cycle as well as the susceptibility of the steel to temper-brittleness and difficulties due to retained austenite should be investigated. One or more of the parts can be tempered with the selected tempering cycle and impact and tensile specimens removed from the part. An inspection of the tensile results will indicate whether the strength meets the prescribed requirements. The reduction of area measurements will indicate whether the steel is of the desired quality; for rolled and forged parts, the difference between the reduction of area for transverse and for longitudinal specimens is, in general, a good index of quality, being less the better the quality.

If notched-bar results that are not completely satisfactory are obtained, tests should be made for temper brittleness and retained austenite effects, unless the design is such that positive elimination of these difficulties can be made. Figures 75 to 77 are useful in determining whether the notched-bar energies are lower than should be expected. Low-temperature notched-bar tests are more sensitive than room temperature tests in detecting poor material.

Usually presence of retained austenite effects can be readily checked by comparing the properties of a sample part given the selected treatment with those of another sample from the same heat given the same treatment but cooled to a lower temperature immediately following the quench (before the temper). This will often mean sub-zero cooling. If the parts are so large that this is not practical even on one sample, the recommended procedure is to cut several similar pieces from similar locations in a sample part after quenching, give some but not all of the pieces additional cooling, then temper all together and compare the notched-bar properties obtained.

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*Such as poor steel quality (porosity, poor inclusion distribution, etc.) temper-brittleness, or high temperature decomposition products from retained austenite.

(preferably at low temperatures) with and without the additional cooling. The amount of austenite retained increases very markedly with increasing carbon content; in low carbon steels (.30% or less) it is not likely to introduce difficulties unless the alloy content is high or the steel is not cooled to room temperature prior to tempering.

The presence of temper-brittleness and the magnitude of the effect may be somewhat more difficult to establish. If the tempering temperature is above about 1100° F. (600° C.), the effect of this transformation may be determined by comparing the notched-bar properties (preferably at low temperatures) with results from specimens from small sections taken from the part, reheated to the chosen temperature held for a short time, and quenched from the temper. If the tempering temperature is below 1100° F. (600° C.), the procedure described in Step 6 is applicable.

If no evidence of incomplete hardening, temper-brittleness, or retained austenite effects can be found and the toughness of the steel of the part is still less than expected, then variables associated with steel making should be investigated. Poor melting practice, for example, may result in low impact properties. Grain boundary precipitates resulting in "conchoidal" fracture may be present (Chapter IV). "Dirty" steel, containing many inclusions, and steel containing flakes or other defects can be expected to have poor properties; these defects should be evident on macro-etching or microscopic examination.

Step 8: Readjustment of Design

If any of the difficulties mentioned above are found, modifications of the design is necessary. If the part has been found to harden incompletely, the alloy content or the severity of the quench may be increased. These modifications may result, however, in an increase in quench-cracking, which may necessitate additional changes.

If the parts are completely hardened but cracking or excessive warping occurs (even when the quench is immediately followed by a low temperature tempering treatment) adjustments are again necessary. If the composition was chosen so as to minimize quench-cracking, little can be done in the way of composition changes unless the hardenability is excessive and may be reduced. (The minimum analysis can be checked by making sample parts of steels of the same type but somewhat lower in alloy or carbon content. Sometimes parts can be made of the same steel but somewhat thicker in the most slowly cooled section.) It may be possible to reduce the carbon content of the steel if the strength can be obtained with a lower (permissible) tempering temperature. It is most important for minimizing quench-cracking and for obtaining the maximum toughness to use the absolute minimum amount of carbon. If the alloy or carbon content is not excessive then a redesign of the quenching procedure or of the shape of the part is called for. A timed quench may be feasible; this ordinarily requires no change in composition. The length of time that the part should be quenched before interruption may be established by quenching a number of sample parts, interrupting the quench after various times, examining the parts for cracks, and after tempering checking their notched bar properties to determine whether the interruption caused deleterious formation of bainite or retention of austenite.

An alternative is to quench a sample part with one or more thermocouples attached to appropriate locations (such as 0.7 the distance from the center to the surface) and determining the time for the average temperature to fall to the estimated M_s or some slightly higher temperature. For moderate or low HD values, a (somewhat low) estimate of the time may be obtained from equation (2-2).

A delayed quench may be employed, and in this case the pearlitic hardenability must be increased. Since delayed quenching is of advantage only if the M_s is brought above the temperature at which cooling becomes exponential, Figures 89 and 90 are useful in judging whether delayed quenching is likely to prove worthwhile. Occasionally, in parts of simple shape over, say, 1 inch thick, increasing the severity of quench will result in a more favorable stress distribution and decrease the tendency to crack. In more complicated parts, changing the quenching severity in recesses by the use of sprays may help.

If none of these measures are successful, the severity of the quenching medium may have to be reduced. (Martempering falls under this heading.) The hardenability will then have to be increased with consequent increased difficulties with retained austenite and temper brittleness. Modification of the part shape, either by masking or by redesign, may be necessary. The fillets may be increased and any sharp changes in section reduced. After any of these changes, several of the preceding steps may have to be repeated, and Step 7 generally must be.

If low toughness has been found due to effects of retained austenite and the alloy and carbon contents were selected to minimize this, then no further changes of composition are possible. It is necessary to alter the heat-treating process. The part may be cooled below room temperature before tempering. A preliminary low-temperature temper to permit transformation of the austenite to a low temperature bainite, before tempering for hardness, may improve the properties. In fact, heating slowly to the tempering temperature may be all that is necessary to transform the retained austenite.

Methods for minimizing temper brittleness have been outlined in Step 6, and need not be further discussed here.

If, in the course of overcoming the various difficulties mentioned, any radical changes in composition, in quenching procedure (other, perhaps, than the introduction of a timed quench or an increase in quenching severity), or in the size of the part are made, most of the steps so far outlined will have to be repeated. Thus, design, quenching practice, and composition, each influence each other. A process of successive approximation is necessary and in difficult cases the steps described herein may have to be gone through several times.

Even when satisfactory properties are obtained in the finished part and quench-cracking and distortion are overcome, metallurgical problems may remain. For example, softening cycles for machinability (which include such possibilities as an ordinary annealing, "cycle-annealing", and low-temperature normalizing followed by tempering) may have to be worked out. Various non-metallurgical troubles may have to be dealt with.

Step 9: Service Tests

Following the final selection of a heat-treatment-composition combination that will produce the desired properties without undue process difficulties, a number of parts should be manufactured. This pilot lot can sometimes be used to establish the product flow and as a final check for manufacturing difficulties. Primarily, however, these parts are intended for service testing. Modification of the shape or a change in strength level may result from such tests. The use of shot-peening or similar pre-stressing method may be indicated. If modifications are necessary, the metallurgical design procedure must be repeated, at least in part.

Step 10: Process and Product Control

The metallurgical designer should cooperate with whoever is responsible for establishing process and product control to see that control procedures are set up which will insure that the metallurgical design is adhered to. Control both of raw steel and of heat-treatment is ordinarily needed. Assurance should be obtained that the steel will be of satisfactory quality, and have the designed hardenability and the designed tempering behavior. (This last is usually controlled by controlling the carbon content and, within broad limits, the alloy content.) Assurance should also be obtained that critical factors of the austenitizing, quenching, and tempering procedures will be held sufficiently close to those selected by the designer.

TABLE VIII

EFFECT OF ALLOYING ELEMENTS IN STEEL

(ON BASIS OF WEIGHT PER CENT)

Effect	Decreasing A_3 Temp.	Decreasing M_s Temp.	Increasing Hardenability Bainitic	Increasing Hardenability Pearlitic	Increasing Temper Brittleness	Increasing A_{cm} Temp.	Carbide Forming Tendency	Increasing Tempered Hardness No Sec. Hard.	Increasing Tempered Hardness With Sec. Hard.
Stronger Positive Effect	C	C	C	C	Mn	Ti	Ti	C	V
	N	Mn	Mn	Mn	Cr	C	V	Mn	C
	Mn	V	P	Mo	Ni	V	Mo	Si	Mo
	Ni	Cr	Cr	P		Mo	W	Mo	Cr
	Zn	Ni	Si	Cr		W (?)	Cr	Cr	Mn
	Cu	Cu	Ni	Si		Cr	Mn	Ni	Si
	Cr (?)	Mo	Cu	Ni		Si		Co	Ni
		W	Mo	Cu		Mn		V (?)	Co
						Ni (?)			
						Cu (?)			
No Effect	Co	Si							
	W	Co	Co (?)	Co	W (?)		Ni		
	Mo	Al			Mo		Si		
	Si								
	V								
	Sn								
	Al								
Stronger Negative Effect	Be								
	P								
	Ti								

X. EXAMPLES

It may be of value to consider an example of the design procedure described in the preceding chapters*. Some of the procedure could undoubtedly be omitted if a reasonable amount of experience is available. However, it is believed that the example will be more illuminating if very little experience is assumed, and the procedure of Chapter IX utilized in full.

Problem:

It is desired to obtain optimum mechanical properties in a steel tube having the following dimensions:

Inside diameter:	3 inches
Outside diameter:	4 inches
Length:	50 inches

There is a series of threads on the outside of the tube, adjacent to one end, but no other notches or abrupt changes in section. A minimum yield strength of 135,000 pounds per square inch (at 0.1 per cent offset) is required. The part is momentarily heated in service to 800° F., but resistance to scaling and corrosion, beyond that obtained with medium-alloy steels, is not necessary.

Solution:

(1). Quenching

An allowance of 1/8 inch on each surface is needed for machining after heat-treatment. The threads can, if necessary, be cut after heat-treatment, but it would be preferable to rough-machine them before heat-treatment. Thus, the part is to be quenched in the form of a long hollow cylinder, 2-3/4 inch I.D. and 4¼ inch O.D., with, if practical, some threads. In accordance with Chapter IX, a water quench is tentatively selected. Suitable equipment for quenching in cold, circulating water is available. To minimize distortion, the part will, tentatively, be quenched vertically. It will be moved up-and-down during the quench to increase the motion of the water relative to the inside and outside surfaces.

(2). Estimation of Required Hardenability

Figures 93 and 100 indicate that the slowest cooling conditions in a hollow cylinder, 2-3/4 inch I.D. and 4¼" O.D., quenched inside and out in still water ($H = 1$ inch^{-1}), are approximately equivalent to the conditions at the center of a 2.3 inch ideal round, and at .31 inch from the quenched end of a Jominy end-quench bar. (Basing the calculation upon

*The data used in this example are hypothetical but are thought to be reasonable.

still water may provide some factor of safety.)

(3). Check of Estimate of Required Hardenability

In order to check this estimate, a sample part is made of a steel that would be expected to half-harden. Table V indicates that SAE 4130 has, on the average, a bainitic hardenability, for half-hardening, estimated at 2.2 inches ideal round, and a higher pearlitic hardenability. This hardenability of 2.2 inches is close to the requirement of 2.3 inches estimated in Step 2. As SAE 4130 steel in the proper size happens to be available, it is used for checking the estimate of required hardenability. A piece of 4130 is machined to the dimensions (before heat-treating) of the part. This sample part is given any convenient austenitizing treatment (in this case, 1 hour hold at 1625° F.) and quenched in the tentatively selected manner until cold*. The quenched piece is sectioned near its midlength, without heating the metal, and a hardness traverse taken. The lowest hardness (near the midwall) is found to be Rockwell C44.

A Jominy end-quench specimen is cut from the same original piece of 4130, adjacent to the material that forms the midwall of the sample part. The specimen is austenitized just as the part was (1 hour hold at 1625° F.), end-quenched, and ground for hardness survey. The hardness survey shows that Rockwell C44 is found at 4/16 or .25 inch from the quenched-end. Thus (to a better degree of approximation than was obtained in Step 2) the cooling conditions in the part, quenched in the tentatively selected manner, are equivalent to .25 inch Jominy distance, or (using Figure 97) to 2.0 inch ideal round.

(4). Selection of Composition

The part must not be affected by heating in service to 800° F. for short periods. One hour at 800° F. corresponds (Figure 125) to only a few minutes at 900° F. If the part were first tempered say, 1½ hours at 900° F., the service conditions, equivalent to only a few additional minutes at this temperature, would produce no appreciable change in hardness difference (Figure 125) and hence in hardness. 1½ hours at 900° F. is equivalent to a hardness difference of 58 Rockwell C. A tempering treatment providing a hardness difference of 58 Rockwell C or greater is, therefore, indicated.

To provide, reproducibly, the required minimum yield strength of 135,000 pounds per square inch it is necessary to aim at a somewhat higher average, such as 145,000 p.s.i. According to Figures 56 and 58, this corresponds to a tensile strength of 159,000 p.s.i., a Brinell hardness of 335, and a Rockwell hardness of C36. The characteristic hardness of the steel should, therefore, be at least 36 + 58 = 94 Rockwell C. If, at a guess (Table VII), the alloying elements will raise the characteristic hardness

*A short length, say 5 inches, of the same inside and outside diameters could have been used instead of a full-length piece, but the latter gives somewhat better reproduction of the quenching conditions that will be obtained in practice.

8 Rockwell C, the carbon content must be sufficient to provide, alone, a characteristic hardness of Rockwell C86. Figure 128 indicates, then, a minimum carbon content of approximately .40 per cent.

It is though advisable to limit the silicon content to an average of .25 per cent. As the steel is to be made in basic open-hearth furnaces, the phosphorus and sulfur contents will probably average .030 per cent each. Because of lack of experience, no "needling agents" will be used. Because of the reportedly greater toughness of fine-grained steels, an ASTM grain size of 7 will be sought.

For full-hardening, Table III then gives the following hardenability factors:

Element	Per cent	Hardenability Factor, Ideal Round	
		Pearlitic	Bainitic
C	.40	.161"	.172"
Si	.25	1.16	1.16
P	.030	1.085	1.085
S	.030	.981	.981

The product of these bainitic hardenability factors is then .212 inches*. Since an ideal round hardenability of 2.0 inch is required, a bainitic hardenability factor of 2.0/.212 = 9.4 remains to be obtained from manganese, chromium, and nickel.

The cost of the steel will be small compared to the cost of the heat-treated part and the carbon content is moderately high, so that quench-cracking might be a problem with water-quenching. It, therefore, appears advisable to select, tentatively, these manganese, chromium, and nickel contents that will minimize the likelihood of quench-cracking. Reference is therefore made to Figures 112 and 113. Figure 113 indicates that for a combined hardenability factor of 9.4, the manganese content should be about .70 per cent. Figure 112 shows the corresponding chromium content to be approximately 1.00 per cent and the nickel content .30 per cent. According to Table III, the pearlitic hardenability factors for these alloy contents are:

Element	Per cent	Pearlitic Factor
Mn	.70	3.87
Cr	1.00	3.33
Ni	.30	1.156

The product of the pearlitic hardenability factors for the elements so far chosen is then 2.96 inches. No further alloy additions are needed to provide the required hardenability of 2.0 inches. To insure the desired fine grain, deoxidation with aluminum is specified.

*The phosphorus and sulfur factors could well have been neglected at this point, and their effect used simply to increase the factor of safety.

Unfortunately, the composition outlined above does not coincide with any of the standard SAE or NE steels. Because of the difficulty and expense of obtaining a special composition in the tonnages involved, it would be preferable to use a standard steel. As an indication of the likelihood of quench-cracking and hence the importance of adhering closely to the composition that will minimize cracking, the M_s temperature of the tentative composition is computed, using Table I:

Element	Per cent		Lowering of M_s (°F.)
C	.40		252
Si	.25		0
Mn	.70		50
Cr	1.00		36
Ni	.30		9
		Total	347

$$\begin{array}{r} 1010 \\ -347 \\ \hline 663 \end{array} = \text{Estimated } M_s$$

With an M_s of the order of 660° F. not much trouble with quench-cracking would be anticipated, particularly if the threads can be cut after heat-treatment. Therefore, consideration of available standard steels is warranted. Reference is made to Table V. These .40 per cent carbon steels whose hardenability for full-hardening (bainitic or pearlitic, whichever is lower) appears to be sufficiently great so that a high percentage of heats would have more than the required 2.0 inch ideal round are:

Steel	Estimated Hardenability Ideal Round (Inches)
2340	2.4
3140	2.7
3141	2.9
3240	2.9
4340	3.1
9840	2.7

The average of the specified composition range for each of these steels is (omitting phosphorus and sulfur):

Steel	C	Mn	Si	Cr	Ni	Mo
2340	.40	.80	.27	—	3.50	—
3140	.40	.80	.27	.65	1.25	—
3141	.40	.80	.27	.80	1.25	—
3240	.41	.50	.27	1.05	1.82	—
4340	.40	.70	.27	.80	1.82	.25
9840	.40	.80	.27	.80	1.00	.25

M_s temperatures for these steels are found in Table VI and the current price extras ascertained, with the following results:

Steel	M_s Temp. °F.	Price Extra (¢/lb.)
2340	600	1.70
3140	650	0.85
3141	650	0.85
3240	640	1.35
4340	630	1.70
9840	650	1.30

Steels 2340, 3240, and 4340 are apparently over-alloyed for the purpose: they have the lowest M_s temperatures and the highest price*. 3141 appears to be unnecessarily higher in alloy content than 3140 and therefore must have a slightly lower M_s. 9840 is appreciably more expensive than 3140. 3140 is readily obtainable at the moment, and appears likely to remain so. It is, therefore, the new tentative choice.

(5). Austenitizing and Tempering Procedures

Using Figure 128 and Table VII, the characteristic hardness of SAE 3140 is estimated:

Element	Per cent	Contribution to Charac. Hard. No Sec. Hard.	Full Sec. Hard.
C	.40	85.5	85.5
Mn	.80	3.2	3.2
Si	.27	1.1	1.1
Cr	.65	1.0	3.2
Ni	1.25	1.2	1.2
	Total	92.0	94.2

The tempering treatment required (Step 4) is equivalent to more than 1 hour at 850° F. and therefore, according to Chapter VIII, practically the full secondary hardening arising from the presence of chromium is to be expected. The characteristic hardness then happens to coincide with the 94 Rockwell C estimated in Step 4, and no readjustment in composition need be made to take care of the hardness response to tempering.

To check the probable susceptibility to temper-brittleness, reference is made to Figure 134. To use this figure, the combined pearlitic hardenability factor for manganese, chromium, and nickel is calculated for SAE 3140:

*The M_s temperatures are, of course, lower the greater the deviation of the composition from that tentatively selected earlier.

Element	Per cent	Pearlitic Factor
Mn	.80	4.28
Cr	.65	2.51
Ni	1.25	1.65
	Product	17.7

According to Figure 134, a steel similar to 3140 but containing .25-.40 molybdenum shows a very slight embrittlement in a test of moderate sensitivity after a severe embrittlement treatment. Without the molybdenum, embrittlement will take place more rapidly.

To check the likelihood of embrittlement, the tempering treatment is estimated. The minimum practical tempering time for a part of the prescribed dimensions in an air furnace (the type most readily available) is considered to be about ½ hour. For the required hardness difference of 58 Rockwell C (Step 4), Figure 125 then indicates a tempering temperature of about 940° F. Such a treatment might cause appreciable embrittlement in a moderately susceptible steel. Pending an experimental check, however, it is assumed that no serious embrittlement will occur in SAE 3140, and this steel is kept as the tentative selection.

There are few numerical data available on which to base an estimate of the lower boundary of the A_1 temperature range for 3140. However, the steel is not particularly high in manganese and nickel and contains some chromium. Probably the lower boundary is above 1250° F.; almost certainly it is above 1225° F. With the tempering treatments under consideration there is no danger of forming austenite during tempering.

SAE 3140 contains only small amounts of carbide-forming elements. A moderate austenitizing treatment should provide homogeneous austenite. For iron-carbon alloys, the equilibrium austenite field, for .40 per cent carbon, begins at about 1475° F.; with the alloy content of 3140 it would begin slightly lower. An austenitizing treatment of 1 hour at 1575° F. appears sufficient.

(6). Experimental Steels

Small pieces for preliminary tests are obtained from several heats of SAE 3140. From the heats in stock, four are selected having high, low, and medium analyses with respect to carbon and alloy, as follows:

Heat No.	Designation	C	Mn	P	S	Si	Cr	Ni	Mo
S4307	A	.42	.77	.028	.021	.32	.69	1.19	.04
F1129	B	.40	.86	.031	.020	.26	.73	1.23	.03
S3879	C	.41	.73	.034	.026	.29	.59	1.12	.05
N7214	D	.38	.75	.027	.018	.25	.62	1.21	.04
Specification		.38 -.43	.70 -.90	.040 max.	.040 max.	.20 -.35	.55 -.75	1.10 -1.40	--- ---

Unfortunately no heats with extreme simultaneous variations of carbon and alloy can be found. The standard practice of the steel makers has been to make their alloy steels "inherently fine-grained".

Design of preliminary machinability heat-treatment is a problem that has not been explicitly considered in this volume. For uniformity and machinability a preliminary treatment is considered desirable. The microstructure that will be obtained with a preliminary normalizing treatment (air-cooling from the austenite range) is, therefore, estimated. The stock for the production part will be in the form of heavy tubing, 4½ inch outside diameter and 1 inch thick. No charts are given in Chapter V for air-cooling of hollow cylinders, but the tubing should be roughly equivalent to a plate of slightly greater thickness, say 1¾ inches. According to Figure 92, this in turn is equivalent, for still air-cooling, to about 18 inches ideal round. As Table V gives only 4.8 inches as the ideal round size for half-hardening SAE 3140 to pearlite, it appears that air-cooling the tubing will produce a structure substantially all pearlite and proeutectoid ferrite. In .40 per cent carbon steels, pearlitic structure produced by continuous cooling are usually soft enough to machine satisfactorily. A tentative preliminary treatment of 1 hour at 1525° F., followed by air-cooling, is, therefore, selected.

The pieces of SAE 3140 available are not in the form of 4½ inch O.D., 1 inch wall tubing, nor can they be machined to this size. As a rough approximation to the selected preliminary treatment, they are austenitized 1 hour at 1525° F. and air-cooled in the sizes in which they happen to be available: bar stock ranging from 1¾ inch cold-drawn hexagon to 2 inch hot-rolled round. (According to Chapter V, these sizes correspond on air-cooling to approximately 13 to 17 inch ideal rounds.) As a check, Brinell hardness measurements are then made of the bar surfaces (after proper surface preparation), with the following results:

Heat	BHN
A	235
B	245
C	215
D	200

These hardnesses indicate that the structure of the air-cooled bars is probably pearlitic and tend to confirm the view that the air-cooled tubes (which are slightly larger in equivalent section and so will cool more slowly) will be soft enough to machine satisfactorily.

Two small pieces cut from a normalized bar of each heat are austenitized for one hour at each of the following temperatures: 1500, 1525, 1550, 1575, 1600, 1650° F. One of the two pieces is then quenched in water and examined metallographically for carbide solution. The other piece is transferred to a furnace at 1325° F., held one hour (to precipitate proeutectoid ferrite at the grain boundaries), and quenched in water. This piece is intended to be examined metallographically for austenitic grain size. The results obtained for carbide solution are:

Temp. °F.	Undissolved Carbides Visible			
	Heat A	Heat B	Heat C	Heat D
1500	Many	Many	Few	Few
1525	Trace	Few	None	None
1550	None	None	None	None
1575	None	None	None	None
1600	None	None	None	None
1650	None	None	None	None

While an austenitizing temperature of 1550° F. apparently provides complete carbide solution, the previously suggested 1575° F. seems desirable to provide greater safety.

To reduce the metallographic work, only the grain-size samples austenitized at 1575° and at the highest temperature (1650° F.) are immediately examined. The results are:

Temp. °F.	A.S.T.M. Austenitic Grain Size			
	Heat A	Heat B	Heat C	Heat D
1575	8	9	7	8
1650	7½	9	7	8

No examination of the other grain size samples is necessary, for the grain size is fine at the tentative austenitizing temperature and remains so to considerably higher temperatures.

A standard Jominy end-quench hardenability specimen is then machined from the normalized bar of each heat, austenitized 1 hour at 1575° F., and end-quenched. Two opposite flats are ground .050 inches deep (to insure removal of decarburization) on each specimen and hardness curves made. The hardness obtained is plotted, for each heat, against the distance from the quenched end, the latter being on a logarithmic scale. The hardness curves drop slowly near the quenched end, then abruptly begin to drop more rapidly, and finally, as the air-cooled end is approached, level out. The beginning of the abrupt drop is:

Heat:	A	B	C	D
Distance, in 1/16 inch:	8	9	6	6½

The lowest of these values, 6/16 inch, is, according to Figure 93, equivalent to 2.4 inches ideal round. The hardenability, therefore, seems to be greater than the necessary 2.0 inches, for compositions fairly well covering the specified range for SAE 3140.

Because of the doubt as to whether temper-brittleness will occur, the tempering behavior of the four heats is checked at this point. Small pieces cut from the normalized bar of each heat are austenitized 1 hour at 1575° F., water-quenched, and tempered ½ hour at each of these temperatures: 700, 900, 1050, 1200° F. After tempering the pieces are quenched and ground. Rockwell C readings are taken on each piece and plotted against the tempering

temperature for each steel. Interpolation on the plots gives the following tempering temperatures for Rockwell 036:

Heat:	A	B	C	D
°F:	910	920	900	890

Apparently the estimate of 940° F. was slightly high.

From the air-cooled bar of each heat, four 5/8 inch round blanks, each 5 inches long, are machined, given the selected austenitizing treatment, and water-quenched. Two of the four blanks are tempered for ½ hour at the temperature just determined as appropriate for the heat, and quenched in water following the temper. The other two blanks are tempered for six minutes in a salt bath and then water-quenched. The temperatures for the six-minute tempering are chosen, with the aid of Figure 125, to give the same hardness in six minutes as is obtained in ½ hour at the tempering temperatures used for that time. These temperatures are:

Heat:	A	B	C	D
°F:	960	970	950	940

From each blank two V-notch Charpy specimens are machined. Two hardness readings are taken on each. One of each pair of specimens is broken at room temperature, the other at -40°, with the following results (hardness readings averaged):

Tempering	Room Temp. Test			-40° Test		
Time (Min.)	Ft.-Lbs.	Fract.	Rockwell C	Ft.-Lbs.	Fract.	Rockwell C
HEAT A						
30	31.7	F	35.5	23.9	F	35.7
	34.0	F	35.5	25.2	F	35.2
6	35.2	F	35.2	29.7	F	35.0
	33.4	F	35.5	28.1	F	35.5
HEAT B						
30	32.7	F	36.0	17.3	Cbf	36.0
	30.9	F	36.2	15.4	Cbf	36.2
6	34.3	F	36.2	26.8	F	36.0
	35.2	F	35.7	24.9	F	36.0
HEAT C						
30	34.6	F	36.7	28.8	F	36.7
	35.2	F	36.5	28.3	F	37.0
6	34.6	F	36.7	30.4	F	36.5
	37.2	F	37.0	29.7	F	36.7

CONTINUED

Tempering Time (Min.)	Room Temp. Test Ft.-Lbs.	Fract.	Rockwell C	-40° Test Ft.-Lbs.	Fract.	Rockwell C
HEAT D						
30	37.5	F	35.5	29.7	F	36.2
	36.0	F	36.0	29.7	F	35.7
6	37.2	F	35.7	30.9	F	36.0
	38.5	F	36.0	32.1	F	36.0

(Fractures: F = fibrous; Cbf = bright crystalline with fibrous border.)

It is evident that the half-hour treatment embrittles heat B, which is highest in alloying elements. Heat A shows some embrittlement, and heats C and D a little.

The part under design is too large to temper with available salt, lead, or induction furnaces, and the quantity to be produced does not warrant purchase of a special tempering furnace. It is, therefore, not practical to reduce the tempering time below ½ hour, and hence to raise the tempering temperature. The composition of the SAE 3140 cannot be changed and the steel still kept standard. Therefore, to avoid temper-brittleness, it is necessary to abandon SAE 3140.

Since the contents of manganese, chromium, and nickel in SAE 3140 are not excessive for the hardenability required, the indicated composition changes are of but two kinds: additions that raise the characteristic hardness (and so the tempering temperature) and molybdenum additions to reduce the rate of embrittlement. While increasing the carbon would raise the characteristic hardness, it would decrease the toughness (of the unembrittled steel) and increase the tendency towards quench-cracking. Additions of manganese, chromium, or nickel (Table VII) would increase the susceptibility to temper-brittleness and increase the likelihood of quench-cracking by lowering the martensite range (Table I). Increase of silicon is not possible if the choice is to be confined to SAE and NE steels. Addition of vanadium would probably be satisfactory, though it would lower the martensite range. Addition of molybdenum would raise the characteristic hardness as well as provide the specific retarding effect of molybdenum upon temper brittleness; it appears to have no significant disadvantages.

The steels considered in Step 4 to be second choices to 3140 were 3141 and 9840. SAE 3141 differs from 3140 only in containing slightly more chromium; its use would not be in accordance with the considerations just mentioned. (This is true also of SAE 3240.) SAE 9840 contains a little more chromium and a little less nickel than 3140, but, more important, contains also .20 - .30 per cent molybdenum. This is a change in the direction indicated to be desirable. SAE 9840 is, therefore, the now tentative choice.

(5). Austenitizing and Tempering Procedures (repeated)

The characteristic hardness of SAE 9840 is estimated from Table VII and Figure 128 as follows:

Element	Per cent		Contribution to Charact. Hard. No Sec. Hard.	Full Sec. Hard
C	.40		85.5	85.5
Mn	.80		3.2	3.2
Si	.27		1.1	1.1
Cr	.80		1.2	4.0
Ni	1.00		1.0	1.0
Mo	.25		.6	5.0
		Total	92.6	99.8

(The chromium and molybdenum contents are sufficiently low that, according to Figure 130, the presence of molybdenum does not decrease the secondary hardening effect due to chromium.) For the desired hardness of Rockwell C36, the characteristic hardnesses of 92⅓ and 100 correspond respectively to hardness differences of 56⅓ and 64 Rockwell C. These are equivalent (Figure 125) to 1 hour tempers at 880° and at 1060° F. While full chromium secondary hardening would develop at these temperatures, molybdenum secondary hardening will not be quite fully developed even at the higher temperature (Chapter VIII). If the characteristic hardness developed is estimated at 98 Rockwell C, the hardness difference will be C62 and the tempering treatment 1 hour at 1010° or ½ hour at 1030° F. The estimate of 98 Rockwell C appears reasonable for such tempering treatments.

It might be thought that since the characteristic hardness and hardness differences are above the minimum to avoid softening in service (Step 3), the carbon content might profitably be reduced. However, the likelihood of temper-brittleness must be checked first. The combined pearlitic hardenability factor for manganese, chromium, and nickel will not be very different from that for SAE 3140, calculated in Step 5 as about 18. Figure 134 indicates that such a steel, containing .25 - .40 molybdenum, shows slight temper brittleness after a severe embrittlement treatment. A tempering treatment of ½ hour at 1030° F. or even 1 hour at 1010° F. probably will not introduce appreciable temper brittleness. However, if the carbon content were reduced and the tempering temperatures lowered correspondingly, embrittlement would be likely to take place. Thus, the carbon content had, at least for the moment, best be left at .40 per cent.

Since NE 9840 contains appreciable molybdenum and chromium and only moderate quantities of manganese and nickel, the lower boundary of its A_1 temperature range is hardly likely to be below 1275° F. Unless the estimate of characteristic hardness is greatly in error, there is thus no danger of austenite formation during tempering.

Because of its greater content of carbide-forming elements, NE 9840 will require a higher austenitizing temperature than SAE 3140. One hour at 1625° F. might be a fair estimate.

(6). Experimental Steels (repeated)

A length of 1½ inch round hot-rolled bar is obtained from each of three selected heats of NE 9840, having the following check analyses:

Heat No.	Designation	C	Mn	P	S	Si	Cr	Ni	Mo
L1431	E	.39	.74	.029	.022	.26	.72	1.05	.23
Q3375	F	.41	.79	.023	.028	.24	.85	.98	.26
R1902	G	.39	.86	.025	.017	.31	.85	1.09	.22
Specification		.38 -.43	.70 -.90	.040 max.	.040 max.	.20 -.35	.70 -.90	.85 -1.15	.20 -.30

According to Table V, the ideal round size for half-hardening NE 9840 to pearlite is 8.9 inches, and to bainite, 4.9 inches. The stock for the part is, on cooling in still air, equivalent to about 18 inches ideal round (Step 5). A preliminary treatment involving air-cooling from the austenitizing temperature will probably result in a structure principally pearlite and proeutectoid ferrite with likely some bainite. Even with a low-temperature austenitizing treatment that leaves undissolved carbides and so increases the speed of pearlite formation on cooling, enough bainite might be present to raise the hardness above that economical for machining. A tempering treatment may, therefore, be advisable after the air-cooling; one hour at 1225 - 1250° F. seems to be as high as would be safe without a check of the low-temperature boundary of the A_1 range. One hour at 1525° F., air-cooling, and tempering one hour at 1225 - 1250° F. is, therefore, tentatively chosen as the preliminary treatment*.

The 1½ inch round bars of heats E, F, and G are austenitized 1 hour at 1525° F., air-cooled to room temperature, and their hardnesses measured. They are then tempered 1 hour at 1225° F. and their hardnesses measured again, with the following results:

Heat	Brinell Hardness Before Temper	Brinell Hardness After Temper
E	245	230
F	260	235
G	305	230

On the small bars, at least, a temper is apparently desirable to achieve a uniform, low, hardness.

Pieces cut from the normalized-and-tempered bars are used, as described in Step 6, to determine the effect of austenitizing temperature upon

*Air-cooling (normalizing) and tempering is commonly more economical than treatments producing pearlite, either by furnace-cooling or by isothermal decomposition of austenite. Full consideration of metallurgical design for machinability lies, however, outside the scope of this volume, and little further attention will be given here to such design.

carbide solution and austenitic grain size, with the following results:

Temp. °F.	Undissolved Carbides Visible			A.S.T.M. Grain Size		
	Heat E	Heat F	Heat G	Heat E	Heat F	Heat G
1550	Many	Many	Many			
1575	Few	Few	Many			
1600	None	Trace	Trace			
1625	None	None	None	7	8	8
1675	None	None	None	7	7½	8

Though no carbides are found in these particular heats at 1625°, they are found at 1600° F. The previously estimated austenitizing temperature of 1625° F. might not provide sufficient margin of safety to take care both of variations among different heats and of discrepancies between the nominal austenitizing temperature and that obtained in production. A treatment of 1 hour at 1650° F. would be decidedly preferable, and, on the basis of the grain size measurements, seems to have no disadvantage.

With this austenitizing treatment, a Jominy end-quench test is made upon the air-cooled and tempered bar of each heat. The abrupt drop in hardness is found to begin at the following distances from the quenched end of the specimen:

Heat:	E	F	G
Distance, in 1/16 inch:	7	9	9

The minimum hardenability, 7/16 inch, is equivalent (Figure 97) to 2.7 inches ideal round, and, therefore, exceeds the required 2.0 inches by a moderate amount. Since heat E is fairly close to the lower limit of the specified chemical range, the great majority of NE 9840 heats should have sufficient hardenability.

The hardness response to tempering is checked by austenitizing several small pieces from each of the heats E, F, and G for the selected 1 hour at 1650° F., quenching them, and tempering a piece from each heat ½ hour at each of these temperatures: 900, 1000, 1100, 1200° F. Rockwell C hardness measurements are then taken on each piece and the hardness plotted versus tempering temperature for each heat. By interpolation, the temperature for Rockwell C36 is found to be:

Heat:	E	F	G
°F.:	1040	1060	1060

With such temperatures, the likelihood of embrittlement at the tempering temperature appears small. It, therefore, seems advisable to check the embrittlement occurring at the longer and more convenient tempering time of 1 hour, rather than at the shorter time of ½ hour. The temperatures for 1 hour temper corresponding to the above temperatures for ½ hour temper are, according to Figure 113:

Heat:	E	F	G
°F.:	1020	1040	1040

Corresponding temperatures for 6 minutes temper are:

Heat:	E	F	G
°F.:	1100	1120	1120

The 1½ inch bars will have to be cut down to about 5/8 inch rounds or ½ inch squares in the process of obtaining Charpy bars, and this might just as well be done before heat treatment, when machining is easier. From the air-cooled and tempered bar of each heat, four 5/8 inch round blanks are machined, austenitized 1 hour at 1650° F., and water-quenched. Two of the four blanks are then given a 1 hour temper and two a 6 minute temper, at the temperatures just mentioned, and quenched in water from these temperatures. Two V-notch Charpy specimens are machined from each blank, and, after hardness readings are taken, one is broken at -40° while the other is broken at room temperature. The results are:

Tempering	Room Temp. Test			-40° Test		
Time (Min.)	Ft.-Lbs.	Fract.	Rockwell C	Ft.-Lbs.	Fract.	Rockwell C
HEAT E						
60	37.2	F	36.0	29.1	F	36.2
	36.0	F	36.2	28.1	F	36.5
6	37.5	F	36.0	30.9	F	36.0
	36.4	F	36.5	30.4	F	36.5
HEAT F						
60	35.2	F	35.7	29.7	F	35.5
	34.6	F	36.2	28.3	F	36.0
6	34.3	F	36.2	28.8	F	35.7
	36.0	F	35.7	28.1	F	36.0
HEAT G						
60	37.5	F	35.7	28.3	F	35.5
	38.5	F	35.2	27.5	F	36.0
6	37.8	F	35.5	32.1	F	35.7
	37.2	F	35.7	31.1	F	35.5

The -40° results indicate a slight embrittlement of Heat G, and probably also of heat E, in the 1 hour treatment. An additional 5/8 inch blank is obtained from each of these heats, austenitized 1 hour at 1650° F., water-quenched, tempered ½ hour, and water-quenched. The tempering temperature is 1040° F. for heat E and 1060° F. for heat G. The two Charpy specimens obtained from each blank are broken at -40° F. with the following results:

Heat	Ft.-Lbs.	Fract.	Rockwell C
E	30.4	F	36.5
	31.1	F	36.0
G	31.5	F	35.7
	30.9	F	36.0

It thus appears that no appreciable temper brittleness occurs during the half-hour tempering treatments, for a V-notch Charpy test at -40° is a very sensitive test at the hardness level involved. It is questionable whether significant embrittlement will take place on cooling a part made of NE 9840 from the tempering temperature, but this is best determined later on a sample part.

Comparison of the room-temperature results with Figure 77 shows that the impact energies obtained are about what might be expected for tempered martensitic .40 per cent carbon steel at Rockwell C36.

(7). Experimental Parts

It is not found possible to obtain small quantities of NE 9840 tubing of the required size for making pilot parts. However, NE 9840 billets from which the parts could be hogged out prove to be available. Several billets are obtained from each of two heats, one high and the other low in carbon and alloying elements, as follows:

Heat No.	Designation	C	Mn	P	S	Si	Cr	Ni	Mo
A4494	H	.42	.83	.024	.019	.31	.86	1.01	.20
M8513	J	.38	.76	.030	.017	.25	.75	.96	.22
Specification		.38 -.43	.70 -.90	.040 max.	.040 max.	.20 -.35	.70 -.90	.85 -1.15	.20 -.30

One billet of each heat is given the tentatively selected machinability treatment of 1 hour at 1525° F., air-cool, and temper 1 hour at 1225° F. Brinell hardnesses are then checked and found to be 225 for heat H and 215 for heat J.

Heat J, on the basis of its composition, would be expected to be lower in hardenability than any of the 9840 heats so far tried. In order to find out whether this heat is likely to have sufficient hardenability to harden in the part, a Jominy end-quench hardenability specimen is machined from the heat J billet, with its midline about 1-3/4 inches from the center*. The specimen is austenitized 1 hour at 1650° F., end-quenched, and a hardness survey made. No abrupt drop in hardness is found up to 6/16 inch from the quenched end, corresponding (Figure 97) to 2.4 inches ideal

*1-3/4 inches is the average radius of the part as heat-treated.

round. The hardenability of heat J, therefore, appears sufficient.

A piece from the normalized-and-tempered billet from each heat is then rough-machined to the dimensions of the parts before heat-treating (including rough-machined threads), but with several inches extra length for test specimens. This extra length is then cut off. One small piece is cut from the extra length of each heat, given the selected austenitizing treatment of 1 hour at 1650° F., and water-quenched. These pieces are then tempered ½ hour at 1050° F. and Rockwell C readings taken on each. Heat H is found to average Rockwell C36.8 while heat J runs about C35.6. According to Figure 125, to lower the hardness of heat H by 0.8 Rockwell C (from C36.8 to the desired C36.0) the tempering temperature should be raised from 1050° to 1070° F., while to raise the hardness of heat J by 0.4 Rockwell C the temperature should be lowered from 1050° to 1040° F.

For comparison with the results to be obtained on the full-size section, enough pieces ½ inch by 3/4 inch (the wall thickness) are cut from the extra length of each sample part to provide 2 tensile and 4 Charpy specimens*. These pieces are austenitized 1 hour at 1650° F., water-quenched, tempered ½ hour, and quenched in water. The tempering temperature is 1070° F. for heat H and 1040° F. for heat J. Tensile and V-notch Charpy specimens are machined from the heat-treated pieces. The Charpy specimens are notched on the side adjacent to the inner cylindrical surface of the part. Two of the Charpy specimens from each heat are broken at -40°, the other two and both tensile specimens at room temperature. The results are:

Heat	Yield Strength (0.1% Offset) Lbs./sq. in.	Tensile Strength Lbs./sq. in.	Elong. %	Red. of Area
H	146,500	161,400	14.6	54.5
	148,400	162,100	15.1	53.9
J	146,200	161,800	16.0	56.3
	144,900	160,300	15.4	57.8

Heat	Notched-Bar Room Temp. Ft.-Lbs.	Room Temp. Fract.	-40° Ft.-Lbs.	-40° Fract.	Hardness Rockwell C	
H	38.5	F	31.1	F	35.7,	36.0
	39.2	F	30.4	F	36.2,	35.7
J	40.3	F	32.6	F	36.2,	36.5
	41.6	F	31.5	F	36.0,	35.7

The two sample parts are austenitized 1 hour at 1650° F. and water-quenched in the equipment and, in the manner to be used for production parts.

*Transverse specimens would be preferable but cannot be obtained from a part of the dimensions involved.

The heat H part is tempered ½ hour at 1070° F. and the heat J part ½ hour at 1040° F. For the production part, air-cooling from the tempering temperature is preferred to quenching, if it does not introduce temper-brittleness, since air-cooling introduces less residual stresses and so less likelihood of distortion on subsequent machining. The sample parts are, therefore, air-cooled from the tempering temperature.

Measurements of the heat-treated parts show no significant distortion during heat treatment. Each part is then sectioned longitudinally, along a diameter, and one-half macroetched. Quenching cracks are found at the base of the threads in the heat H part, but not in the heat J part.

From the midwall of the other half of each sectioned part, at approximately the midlength, two tensile and 4 V-notch Charpy longitudinal specimens are machined. The Charpy specimens are notched on the side adjacent to the inner cylindrical surface of the part, just as were the Charpy specimens heat-treated as small pieces. Two of the Charpy specimens from each heat are tested at -40°, the other two and both tensile specimens at room temperature, with the following results:

Heat	Yield Strength (0.1% Offset) Lbs./sq. in.	Tensile Strength Lbs./sq. in.	Elong. %	Red. of Area
H	144,900	160,500	15.0	55.2
	144,900	158,700	15.2	55.8
J	142,700	157,400	15.8	56.6
	143,800	158,900	14.7	55.6

Heat	Notched-Bar Room Temp. Ft.-Lbs.	Room Temp. Fract.	-40° Ft.-Lbs.	-40° Fract.	Hardness Rockwell C
H	39.5	F	30.0	F	35.5, 35.7
	40.0	F	30.4	F	35.2, 35.5
J	42.3	F	31.8	F	35.7, 36.2
	40.3	F	32.6	F	35.5, 35.7

The yield-tensile ratios, reductions of area, and notched-bar energies found in the full size section are as high as those found in the pieces treated in small sections. Moreover, the values are within the normal range for .40 per cent carbon tempered martensitic steel of the strength level obtained (Figures 56, 59, 60, and 77). Thus, the choice of steel, the austenitizing treatment, the quenching procedure, and the tempering treatment are apparently satisfactory, except that quench-cracking remains to be eliminated.

(8). Readjustment of Design

Quench-cracking could almost certainly be prevented by machining the threads only after heat-treatment. As it is preferable not to do this,

those changes in quenching practice that can be made without changing the composition are investigated. The simplest of such changes are interrupting the quench, and delaying the quench down to a temperature of perhaps 1250° F. Quenching cannot be delayed to much lower temperatures because the pearlitic hardenability of the steel (estimated in Table V as 6.6 inches ideal round for full-hardening) is obviously insufficient for the part to harden in air with respect to pearlite*.

The thickness of the part wall, as heat-treated, is about 3/4 inch, and the quenching severity (Table II) probably of the order of 1.5 $inch^{-1}$. Thus, the HD product is of the order of 3/4 x 1.5, that is, of the order of 1. With such a relatively low HD product, the whole part will, according to Figure 90, begin to cool exponentially before the surface enters the martensite range, even with quenching temperatures as low as 1250° F. (fractional temperatures of the M_s as high as (650-60)/(1250-60) = .50). Therefore, according to Chapter VII, delayed quenching cannot be expected to offer appreciable mitigation of the quenching stresses.

Since the part thickness is relatively uniform, interrupted quenching would seem to offer promise. The length of time that the part should be left in the quenching tank before interruption may be estimated by equation (2-2). As it was found in Step 3 that the section is equivalent to about 2.0 inches ideal round rather than to the 2.4 inches indicated by Figure 93, the severity of quench is probably somewhat higher than the 1.0 $inch^{-1}$ upon which Figure 93 is based. Table II indicates a severity of the order of 1.5 $inch^{-1}$ for a good water quench. The part can be considered, approximately, as a plate 3/4 inch thick. The average temperature sought may be taken as 700° F. and the water temperature as 60° F. Using equation (2-2) and the definition of c that follows it:

$$H = 1.5 \text{ inch}^{-1}$$

$$D = 3/4 \text{ inch} = .75 \text{ inch}$$

$$a = .009 \text{ inch}^2/\text{sec.} \quad \text{(From Chapter V.)}$$

$$c = HD\left(\frac{4a}{D^2}\right) = 4aH/D = 4(.009)\ (1.5)\ /.75 = .072 \text{ sec.}^{-1}.$$

$$U = \frac{700 - 60}{1650 - 60} = \frac{640}{1590} = .402$$

$$\ln U = -\ln\ (1/.402) = -\ln 2.483 = -.908$$

$$t = -(\ln U)/c = .908/.072 = 12.6 \text{ seconds}$$

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*No graphs showing the ideal round size equivalent to air-cooled hollow cylinders have been provided. However, the 3/4 inch thick wall of the cylinder would be equivalent to slightly more than a 3/4 inch plate. According to Figure 92, the latter, cooled in still air, is equivalent to about 14 inches ideal round.

Equation (2-2) underestimates the time for moderate HD values, so the time required will be probably somewhat greater than 12.6 seconds. Even if the equation and the other approximations introduce an error of a factor of 2, the time required is not over 25 seconds. It is probably not under 10 seconds. For reasonable reproducibility, the time in the quenching medium would have to be controlled within perhaps 20%: 2 seconds in 10 or 5 seconds in 25. Some difficulty is anticipated in maintaining such process control on a production basis with the manually-controlled quenching equipment available, and installation of automatic control equipment is not considered worthwhile. Other possibilities for avoiding quench-cracking are, therefore, considered.

Since the section to be quenched is less than 1 inch thick, it does not appear likely that increasing the severity of quench will, by any chance, result in a stress distribution less conducive to cracking. The only remaining type of quenching practice that could possibly be used for reducing cracking without close control and without changing the composition is stepped quenching in salt or lead. However, neither a salt nor a lead pot of the necessary size is available.

Since heat J, on the low side of the composition range, did not quench-crack, the possibility of selecting low-side heats within the NE 9840 range suggests itself. This is found not to be practical. Lowering the carbon content, while adhering to the selected alloy content is next considered, since NE 9837 and NE 9835 appear on the NE list. However, any lowering of the carbon content will require a corresponding lowering of the tempering temperature to obtain the required strength. Such lowering of the tempering temperature, in the range involved, will lower the contribution of the molybdenum to secondary hardening and hence necessitates some additional decrease in tempering temperature. Since the preliminary tests on heats E and G indicated that the margin by which temper-brittleness is avoided in the selected tempering treatment is small, lowering the carbon content appears inadvisable.

If the type of steel is changed, the choice must be limited to SAE and NE steels. Steps 4 to 6 have previously eliminated all .40 per cent carbon steels except 2340 and 4340. These have lower estimated M_s temperatures than 9840 and so should be even more susceptible to quench-cracking, while their compositions do not seem to fit them for any advantageous types of quenching practice for which 9840 is unsuited.

Steels with more than .40 per cent carbon would have lower toughness in service and a greater tendency to crack on quenching. Their only advantage might be that some could perhaps be hardened with an oil quench, thus greatly reducing the likelihood of quench-cracking. Figure 94, however, indicates that an ideal round size of about 3.8 inches would be required. None of the SAE or NE steels show promise (Table V) of exceeding this value by a safe margin, for full-hardening.

If the carbon content were lowered (retaining a water quench), the only SAE or NE steels that seem to have sufficient hardenability (Table V) are 3135 and the 9800 series. The latter have already been ruled out. 3140 was found to be unsuitable because of temper-brittleness,

and 3135, because of its lower tempering temperature, would be even worse in this respect.

The only remaining alternatives appear to be to mask the threads during quenching, or to machine them after heat-treatment. Masking will appreciably retard the heat flow from the masked section. The threaded section is as thick as any other on the part, and extends sufficiently far from the end so that cooling from the end will not be appreciable over much of it. It does not seem advisable to reduce the safety margin on hardenability, so a steel with higher hardenability would have to be used. SAE 4340 is the only one with any promise (Step 4). This differs from 9840 primarily in having a higher nickel content. The increase in nickel will increase the susceptibility to temper-brittleness and will raise the tempering temperature only very slightly. The use of 4340, therefore, seems likely to introduce temper-brittleness. To establish whether machining the threads after heat-treatment will eliminate quench-cracking, a second piece from the normalized-and-tempered billet of heat H is machined to the dimensions in which the part is to be heat-treated, but is not threaded. This piece is austenitized 1 hour at 1650° F. and water-quenched in the established manner directly to room temperature. To facilitate detection of cracks, the piece is then tempered and sectioned longitudinally. One half of the sectioned part is macroetched; no cracks are found.

The choice then resolves to either machining the threads after heat-treatment or designing an interrupted-quench procedure and establishing close controls to insure its maintenance. After consultation with those responsible for machining, heat-treatment, and process control, it is decided to use a direct quench and machine the threads after heat-treatment.

The design for hardening, then is as follows:

Steel: NE 9840

Part Remarks: Threads to be machined after heat-treatment.

Austenitizing: 1 hour at 1650° F.

Quench: Cold circulating water to room temperature, part vertical, moved up-and-down. (Details of equipment, water temperature, etc., specified.)

Temper: ½ hour at a temperature to give Rockwell C36.

Cool from temper: Air.

(9). Service Tests

A heat of NE 9840 tubing is obtained for manufacture of a pilot lot of parts. The air-cooling and tempering procedure proposed for the machinability heat-treatment is found to be satisfactory. Check of the hardness after air-cooling but prior to tempering indicates that the tempering cannot be omitted. The lot is used to check the proposed machining

set-up and is heat-treated in the selected manner. Checks are made of the hardness of the heat-treated parts, and tensile and Charpy specimens are obtained from a few. Satisfactory results are obtained in these tests. The parts are finished, and service tests made on a number of them. Here too the results are satisfactory.

(10). Process and Product Control

In establishing process controls, adherence to the chemical range specified for NE 9840 should of course be checked. Off-analysis heats may be accepted on a test basis to determine whether the permissible range may safely be widened. Assurance that the steel is "inherently fine-grained" may be obtained by analysis for "metallic" or "soluble" aluminum (aluminum soluble in hydrochloric acid).

It has not been definitely established that all heats falling within the chemical limits will have sufficient hardenability. A requirement might, therefore, be set up that the steel (after the preliminary treatment of 1 hour at 1525° F. air-cooling, and tempering 1 hour at 1225° F.) show no sharp drop in hardness within 4/16 inch from the end of a Jominy end-quench bar austenitized 1 hour at 1650° F.* Heats falling below the specified hardenability might be accepted on a test basis only. If, when a number of production heats had been tested, none were found to fall below the specified hardenability, the hardenability test could be eliminated.

Controls should also be set up to insure that steel containing excessive inclusions or segregation is not used.

The tendency of the steel towards temper-brittleness can perhaps be controlled sufficiently well by Charpy tests on the heat-treated parts, though tests like these made in Step 6 could be used for control.

Check of the hardness of the steel after the machinability heat-treatment would probably provide sufficient control of this heat-treatment.

Some control of the austenitizing temperature, and perhaps a rough control of the time, would be advisable. Close attention would have to be paid to adherence to the selected quenching procedure, particularly temperature of the water, circulation of the water, and motion of the part.

Reference to Table VII indicates that the characteristic hardness may be expected to vary perhaps 6 Rockwell C within the prescribed chemical limits. Since closer control of the tempering response of the steel is not practical, the tempering temperature will have to be varied from heat to

*If the "sharp drop" requirement is not practical, it is possible, though less desirable metallurgically, to set up a definite minimum hardness for 4/16. Figure 114 indicates Rockwell C59 for fully hardened .38 per cent carbon steel. A minimum of C57 might be, therefore, set up if justified by the results on heats E to J.

heat. Control of the tempering cycle can be obtained by hardness checks on the tempered steel, supplemented by overall checks of the process. Such overall metallurgical checks would consist of occasional tensile and Charpy tests on the heat-treated steel.

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