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MATERIALS RESEARCH AND DEVELOPMENT



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THERMOMECHANICAL  
TREATMENTS APPLIED TO  
ULTRAHIGH-STRENGTH BAINITES

10 by

15 D. Kalish and S. A. Kulin,  
Contract N600/19/59811

9 Final Technical Report

11 April 1964

Submitted to:

Metals Branch of the  
Materials Division  
Bureau of Naval Weapons  
Department of the Navy

AI

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MATERIALS RESEARCH AND DEVELOPMENT

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ABSTRACT

The mechanical properties of H11 and AISI 4350 steels were determined for a series of bainites formed from both deformed and undeformed austenites. These bainites were also subjected to subsequent straining and tempering treatments. Similar thermomechanical treatments involving martensite as the essential transformation product were included for comparison purposes. ←

Fifty percent austenite deformation in H11 results in considerable strengthening of the subsequently formed bainite. H11 in the bainitic condition (formed from either deformed or undeformed austenite) is strongly affected by refrigeration in liquid nitrogen (due to conversion of retained austenite) when tempering temperatures below 1000°F are employed. In both cases, a large secondary strengthening is observed on tempering between 1000 and 1100°F. This results in higher strengths and equivalent ductilities as compared to the correspondingly treated martensites. Austenite deformation in 4350 has little effect on the strength of the subsequently formed bainite, although such deformation does enhance the strength of martensite.

H11 and 4350 bainites and martensites exhibit a large response to strain-tempering treatments. For both steels, small deformations (up to 3%) produce large strengthening effects. Yield strengths close to 400,000 psi have been obtained by the application of strain-tempering treatments (50% deformation) to both H11 martensite and bainite. Increasing the carbon content in 43XX steels enhances the strengthening

- a -

response to strain-tempering.

The increments in strength developed by deformation processes both before and after transformation of either bainite or martensite are found to be approximately additive when the individual processes are combined.

Retained austenite determinations were made in H11 which had been subjected to various thermal and thermomechanical treatments. It was found that up to 35 percent austenite is retained when bainite is formed at 550°F from either deformed or undeformed austenite. However, the deformation of the austenite prior to the bainitic transformation produces a retained austenite with decreased stability relative to subsequent refrigeration treatments. In addition, deformation of retained austenite, in a previously undeformed bainitic structure, makes the austenite less stable to further tempering. For tempering temperatures below 1000°F, retained austenite has a considerable effect on the yield strength of the mixed structure. However, the influence of retained austenite on the yield strength is not directly proportion to its volume percent.

The increments in strength developed by deformation processes both before and after transformation to either bainite or martensite are found to be approximately additive when the individual processes are combined.

## Table of Contents

	Page
Abstract	
I. INTRODUCTION . . . . .	1
II. MATERIALS AND EXPERIMENTAL PROCEDURES . . .	3
A. Materials . . . . .	3
B. Designation of Treatments . . . . .	4
C. Thermal Treatments . . . . .	4
D. Deformation of Austenite . . . . .	7
E. Strain-Tempering Martensite and Bainite . . . . .	8
F. Combinations of Thermomechanical Treatments . .	8
G. Refrigeration Treatments . . . . .	8
H. Mechanical Testing . . . . .	9
I. Metallographic Examination . . . . .	9
J. X-Ray Diffraction . . . . .	11
III. RESULTS AND DISCUSSION . . . . .	12
A. Transformation Characteristics of H11 and 4350 Steels . . . . .	12
B. Mechanical Properties of H11 Steel Subjected to Thermomechanical Treatments . . . . .	18
1. Bainite and Martensite - No Deformation (Processes B and M) . . . . .	18
2. Austenite Deformation Prior to Transformation (Processes AB and AM) . . . . .	33
3. Deformation of Bainite (Process BS) . . . . .	45
C. Mechanical Properties of 4350 Steel Subjected to Thermomechanical Treatments . . . . .	56
1. Bainite - No Deformation (Process B) . . . . .	56

Table of Contents (Continued)

	Page
III. RESULTS AND DISCUSSION (Continued)	
2. Austenite Deformation Prior to Transformation (Processes AB and AM) . . . . .	59
3. Deformation of Bainite and Martensite (Processes BS and MS) . . . . .	62
4. Combinations of Thermomechanical Treatments (Processes ABS and AMS) . . . . .	68
5. Effect of Carbon Content on Process MS . . . . .	70
D. General Comparisons and Conclusions . . . . .	70
1. General Comparisons . . . . .	70
2. Conclusions . . . . .	74
References . . . . .	77

## List of Figures

Figure No.		Page
1	Schematic Representation of Thermal and Thermomechanical Treatments.....	5
2	Schematic Representation of Thermomechanical Treatments.....	6
3	Isothermal Transformation Diagrams for (a) H11 and (b) 4350.....	13
4	Light Micrographs of H11 Bainite Transformed (4 hrs.) at (a) 500°F, (b) 550°F, (c) 600°F and (d) 650°F.....	14
5	Electron Micrographs of H11 Bainite Transformed (4 hrs.) at (a) 500°F, (b) 550°F, (c) 600°F and (d) 650°F Plastic Replicas Shadowed.....	15
6	Electron Micrograph of 4350 Bainite Transformed at 550°F (4 hrs.) Carbon Replica Shadowed.....	17
7	Tensile Properties of H11 as a Function of Bainite Reaction Temperature ( $T^b$ ) for Process B. ( $T^a=1850^\circ\text{F}$ , $t_f=4$ hours, $T^q=RT$ , $T^t=550^\circ\text{F}$ ).....	19
8	Mechanical Properties of H11 as Function of Tempering Temperature ( $T^t$ for Process B, $T^a=1850^\circ\text{F}$ , $T^b=550^\circ\text{F}$ , $t_f=4$ hours, $T^q=RT$ or LN).....	21
9	Mechanical Properties of H11 as a Function of Tempering Temperature ( $T^t$ ) for Process M. ( $T^a=1850^\circ\text{F}$ , $T^q=RT$ or LN).....	22
10	Mechanical Properties of H11 as a Function of Isothermal Transformation Time ( $t_i$ ) for Process B. ( $T^a=1850^\circ\text{F}$ , $T^b=550^\circ\text{F}$ , $T^t=550^\circ\text{F}$ , $T^q=RT$ or LN).....	24

List of Figures (Continued)

Figure No.		Page
11	Light Micrograph of H11 Isothermally Held at 550°F for 15 Minutes and Quenched To Room Temperature.....	25
12	Electron Micrographs of H11 for Process B Tempered at 550°F. Plastic Replicas, Shadowed.....	27
13	Yield Strength and Retained Austenite Content of H11 as a Function of Tempering Temperature for Process B. ( $T^a=1850^\circ\text{F}$ , $T^b=550^\circ\text{F}$ , $t = 4$ hours, $T^q = \text{RT or LN}$ ).....	29
14	Effect of Refrigeration Treatment (LN) on Yield Strength and Retained Austenite Content of H11 Subjected to Process B. ( $T^a=1850^\circ\text{F}$ , $T^b=550^\circ\text{F}$ , $t = 4$ hours, $T^q = \text{RT or LN}$ ).....	30
15	Yield Strength H11 as a Function of Percent Retained Austenite for Various Tempering Temperatures in Process B ( $T^a=1850^\circ\text{F}$ , $T^b=550^\circ\text{F}$ , $t_f=4$ hours, $T^q = \text{RT or LN}$ ).....	32
16	Progress of Martensitic Reaction in H11 for Deformed and Undeformed Austenite. ( $T^a=1850^\circ\text{F}$ , $T^d=900^\circ\text{F}$ , Deformation = 50%).....	34
17	Tensile Properties of H11 as a Function Bainite Reaction Temperature ( $T^b$ ) for Process AB. ( $T^a=1850^\circ\text{F}$ , $T^d=900^\circ\text{F}$ , Deformation=50%, $t_f=4$ hours $T^q = \text{RT or LN}$ , $T^t=550^\circ\text{F}$ ).....	35
18	Mechanical Properties of H11 as a Function of Tempering Temperature ( $T^t$ ) for Process AB. ( $T^a=1850^\circ\text{F}$ , $T^d=900^\circ\text{F}$ , Deformation=50%, $T^b=550^\circ\text{F}$ , $t = 4$ hours, $T^q = \text{RT or LN}$ ).....	36

List of Figures (Continued)

Figure No.		Page
19	Tensile Properties of H11 as a Function of Tempering Temperature, ( $T^t$ ) for Process AM. ( $T^a=1850^\circ\text{F}$ , $T^d=1000^\circ\text{F}$ , Deformation = 50%, $T^q=\text{RT}$ ). Courtesy of Pratt and Whitney Aircraft, United Aircraft Corporation.....	38
20	Mechanical Properties of H11 as a Function of Quench Temperature ( $T^q$ ) for Process AB. ( $T^a=1850^\circ\text{F}$ , $T^d=900^\circ\text{F}$ , Deformation=50%, $T^b=550^\circ\text{F}$ , $t_f=4$ hours $T^t=550^\circ\text{F}$ ).....	39
21	Mechanical Properties of H11 as a Function of Isothermal Transformation Time ( $t_i$ ) for Process AB. ( $T^a=1850^\circ\text{F}$ , $T^d=900^\circ\text{F}$ , Deformation = 50%, $T^b=550^\circ\text{F}$ , $T^q=\text{RT or LN}$ , $T^t=550^\circ\text{F}$ ).....	41
22	Electron Micrographs of H11 for Process AB Tempered at $550^\circ\text{F}$ . Plastic Replicas Shadowed.....	42
23	Yield Strength and Retained Austenite Content of H11 as Function of Tempering Temperature for Process AB. ( $T^a=1850^\circ\text{F}$ , $T^d=900^\circ\text{F}$ , Deformation 50%, $T^q=550^\circ\text{F}$ , $t_f=4$ hours, $T = \text{RT or LN}$ ).....	44
24	Effect of Refrigeration Treatment (LN) on Yield Strength and Retained Austenite Content of H11 Subjected to Process AB. ( $T^a=1850^\circ\text{F}$ $T^b=900^\circ\text{F}$ Deformation 50% $T^b=550^\circ\text{F}$ $t_f=4$ hours $T^q=\text{RT or LN}$ ).....	46
25	Yield Strength of H11 as a Function of Percent Retained Austenite for Various Tempering Temperatures in Process AB. ( $T^a=1850^\circ\text{F}$ , $T^d=900^\circ\text{F}$ Deformation = 50% $T^b=550^\circ\text{F}$ , $t_f=4$ hours, $T^q=\text{RT or LN}$ )....	47
26	Mechanical Properties of H11 as a Function of Percent Reduction of Thickness for Process BS. ( $T^a=1850^\circ\text{F}$ $T^b=550^\circ\text{F}$ , $t_f=4$ hours, $T^q=\text{RT}$ , $T^{t1}=550^\circ\text{F}$ , $T^{t2}=\text{RT or } 550^\circ\text{F}$ ).....	49

List of Figures (Continued)

Figure No.		Page
27	Light Micrographs of H11 Bainite Transformed to Bainite at 550°F (4 hrs.) and (a) Tempered at 550°F, (b) Pretempered at 550°F and Strained 10% and (c) Pretempered at 550°F, Strained 10% and Retempered at 550°F.....	50
28	Mechanical Properties of H11 as a Function of Pretempering Temperature ( $T^{t1}$ ) for Process BS. ( $T^a=1850^\circ\text{F}$ $T^b=550^\circ\text{F}$ , $t=4$ hours, $T^q=\text{RT}$ , Deformation = 10% or 35% $T^{t2}=550^\circ\text{F}$ ).....	52
29	Mechanical Properties of H11 as a Function of Retempering Temperature ( $T^{t2}$ ) for Process BS. ( $T^a=1850^\circ\text{F}$ , $T^b=550^\circ\text{F}$ , $t_f=4$ hours, $T^q=\text{RT}$ , $T^{t1}=550^\circ\text{F}$ , Deformation= 10% or 35%).....	53
30	Comparison of Tensile Properties of H11 Bainite Subjected to Various Thermomechanical Treatments.....	55
31	Comparison of Yield Strength of H11 Martensite Subjected to Various Thermomechanical Treatments.....	57
32	Mechanical Properties of 4350 as a Function of Bainite Reaction Temperature For Process B ( $T^a=1550^\circ\text{F}$ , $t_f=4$ hours $T^t=550^\circ\text{F}$ ).....	58
33	Mechanical Properties of 4350 as a Function of Tempering Temperature for Process B. ( $T^a=1550^\circ\text{F}$ , $T^b=550^\circ\text{F}$ , $t_f=4$ hours, $T^q=\text{RT}$ ).....	60
34	Tensile Properties of 4350 as a Function of Tempering Temperature ( $T^t$ ) for Process M and AM. ( $T^a=1550^\circ\text{F}$ $T^d=1050^\circ\text{F}$ Deformation=5%, $T^q=\text{RT}$ ).....	61

List of Figures (Continued)

Figure No.		Page
35	Mechanical Properties of 4350 as a Function of Bainite Reaction Temperature ( $T^b$ ) for Process BS. ( $T^a=1550^{\circ}\text{F}$ , $t_f = 4$ hours, $T^q=RT$ , $T^{t1}=550^{\circ}\text{F}$ , Deformation = 10%, $T^{t2}=550^{\circ}\text{F}$ ).....	63
36	Tensile Properties of 4350 as a Function of Pretempering Temperature ( $T^{t1}$ ) for Process BS. ( $T^a=1850^{\circ}\text{F}$ , $T^b=550^{\circ}\text{F}$ , $t_f = 4$ hours, $T^q = RT$ , Deformation = 10%, $T^{t2}=550^{\circ}\text{F}$ ).....	65
37	Tensile Properties of 4350 as a Function of Retempering Temperature ( $T^{t2}$ ) for Process BS. ( $T^a=1550^{\circ}\text{F}$ , $T^b=550^{\circ}\text{F}$ , $t_f=4$ hours, $T^q=RT$ , $T^{t1}=550^{\circ}\text{F}$ , Deformation = 10%).....	66
38	Yield Strength of 4350 and H11 as a Function of Percent Deformation for Processes BS and MS.....	67
39	Comparison of Yield Strength at 4350 Martensite Subjected to Various Thermo-mechanical Treatments.....	69
40	Influence of Carbon Content on Strain Tempering Response of 43XX Martensites. ( $T^{t1} = 400^{\circ}\text{F}$ , Deformation = 10%) Data for 0.12 and 0.20 Carbon from Reference (8)..	71
41	Comparison of the Response of H11 Bainite and Martensite to Various Thermal and Thermomechanical Treatments.....	72
42	Comparison of the Response of 4350 Bainite and Martensite to Various Thermal and Thermomechanical Treatments.....	73

List of Tables

Table No.		Page
1	Chemical Compositions.....	3
2	M <sub>s</sub> Temperatures of the Steels.....	16
3	Effect of Austenitizing Temperature on the Tensile Properties of 4350 for Processes M and AM.....	62

## I. INTRODUCTION

Significant improvements in the strengths of low alloy steels have been achieved by the application of mechanical deformation either prior to or after the formation of martensite. Ausforming involves the deformation of metastable austenite prior to transformation to martensite and has been shown to develop yield strengths over 300,000 psi in a number of commercial and experimental alloys. Strain-tempering, which involves straining and subsequent tempering of as-quenched or tempered martensite, also develops exceptionally high strengths in low alloy steels.

Both ausforming and strain aging treatments have been applied to Hill steel (1-3) and AISI 4340 steel (4, 5). In general, the reported ductility and fracture toughness values at these high strength levels have been found equal to or less than those developed by the conventional heat treatment.

High strength is a desirable but insufficient requirement since the ductility and fracture toughness characteristics are also important. For ultra-high strength applications, thermal mechanical treatments that result in martensitic structures as the final product do not offer much promise. If strength levels of 300,000 psi or greater are to be used, it will be necessary either a) to improve fabrication and inspection techniques so as to eliminate all flaws that could cause premature failure or b) to develop materials that possess sufficient ductility and toughness at these ultra-high strength levels.

In the search for methods to improve the strength, and possibly the ductility and fracture toughness of low-alloy steels, the utilization of bainitic structures in conjunction with thermomechanical treatments should not be overlooked.

Accordingly, the primary objective of the present investigation is to evaluate the response of certain low-alloy steels to thermomechanical treatments involving transformation to bainites. These processes involve deformation of the steels either before or after the bainitic reaction. The parameters measured are strength, ductility and fracture toughness. Structural changes produced by these thermomechanical treatments are followed by means of light and electron microscopy, and are related to the mechanical property changes.

## II. MATERIALS AND EXPERIMENTAL PROCEDURES

### A. Materials

The two steels H11 and 4350 studied in this investigation are listed in Table 1 and were selected on the basis of the following considerations: a) the compositions of these steels are such that they may be deformed in the metastable austenite condition and then transformed to bainite, b) the tempering behaviors of these steels in the conventionally heat treated condition are well known, and c) the influence of thermomechanical treatments involving the martensitic reaction in these steels has previously been studied<sup>(1-5)</sup>. The results obtained in a preliminary investigation of H11 suggested that further study of this steel would be fruitful. Although 4340 steel meets the above requirements, it was decided that AISI 4350 would be more desirable because the additional carbon provides a lower  $M_s$  temperature and therefore a wider range of high-strength bainites can be achieved.

Table 1  
Chemical Compositions

	<u>Element %</u>								
<u>Steel</u>	<u>C</u>	<u>Mn</u>	<u>Si</u>	<u>Cr</u>	<u>Mo</u>	<u>Ni</u>	<u>V</u>	<u>P</u>	<u>S</u>
H11	0.39	0.25	1.00	5.25	1.39	--	0.54	0.111	0.007
4350	0.52	0.65	0.22	0.88	0.26	1.82	--	0.008	0.006

The steels were obtained as consumable-electrode, vacuum-melted 2-inch wide strip in two thicknesses, 0.250 and 0.140 inch. The thicker stock was used for experiments involving deformations of more than 20% reduction.

### B. Designation of Treatments

Eight basic thermal or thermomechanical treatments were employed in this investigation, and are shown schematically in Figs. 1 and 2. A list of symbols, most of which appear in these figures and are used throughout this report, are defined as follows:

- $T^A$  = austenitizing temperature
- $T^B$  = bainite-reaction temperature
- $T^D$  = austenite-deformation temperature
- $T^Q$  = quench temperature
- RT = room temperature
- LN = liquid nitrogen
- HQ = hot quenched
- OQ = oil quenched
- AC = air cooled
- $t_f$  = time for bainite formation at  $T^B$
- $T^T$  = tempering temperature
- $T^{T1}$  = pretempering temperature (to be followed by deformation of either martensite or bainite)
- $G^{T2}$  = retempering temperature (following deformation of either martensite or bainite)

### C. Thermal Treatments

Processes B and M (Fig. 1) refer to the formation of bainite or martensite, respectively, without any deformation being introduced during the heat treatment cycle. Bainite was formed by austenitizing for 30 minutes in salt at some temperature ( $T^A$ ), hot quenching into salt at the appropriate bainite-reaction temperature ( $T^B$ ), holding to the end of

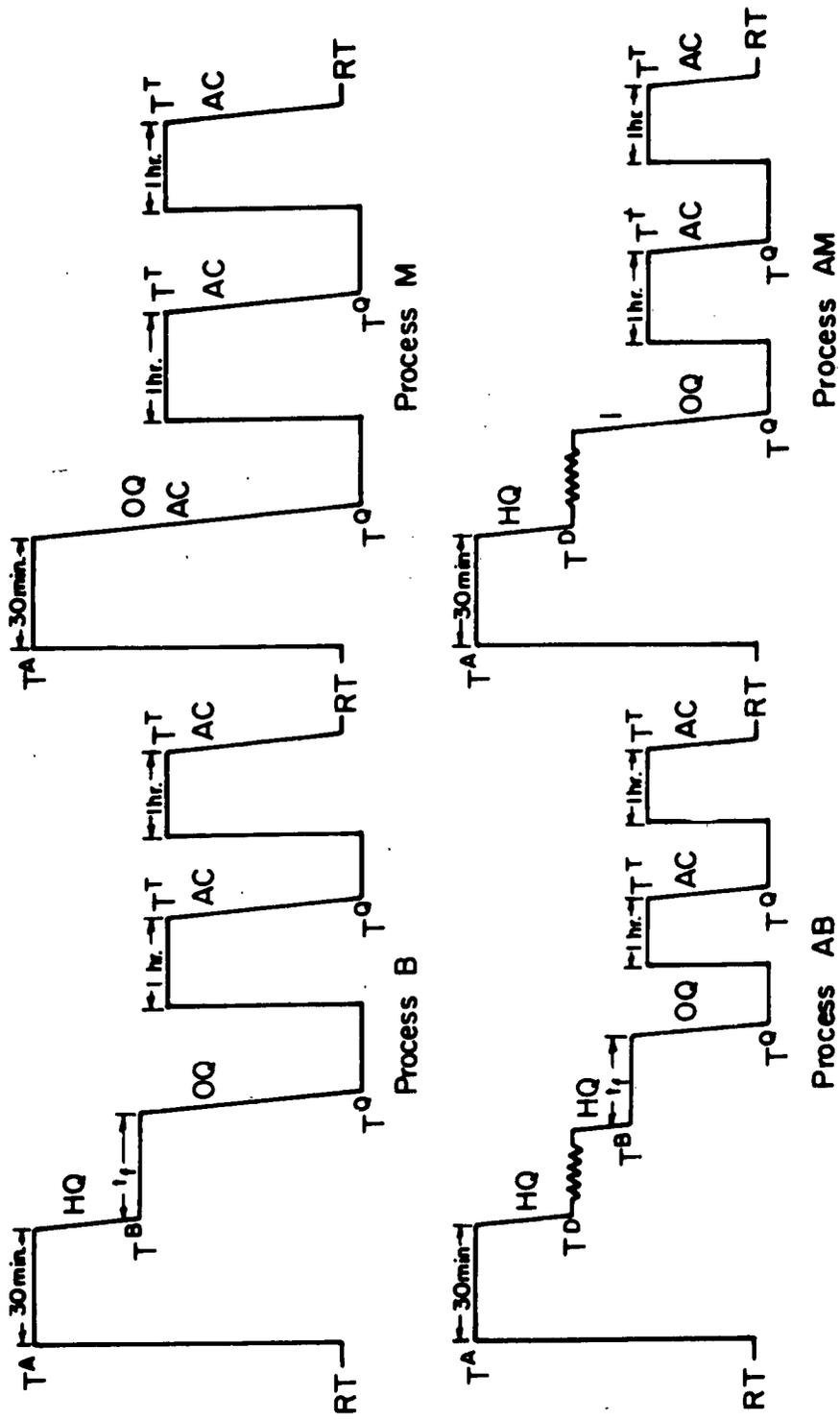


Fig. 1 - Schematic Representation of Thermal and Thermomechanical Treatments.

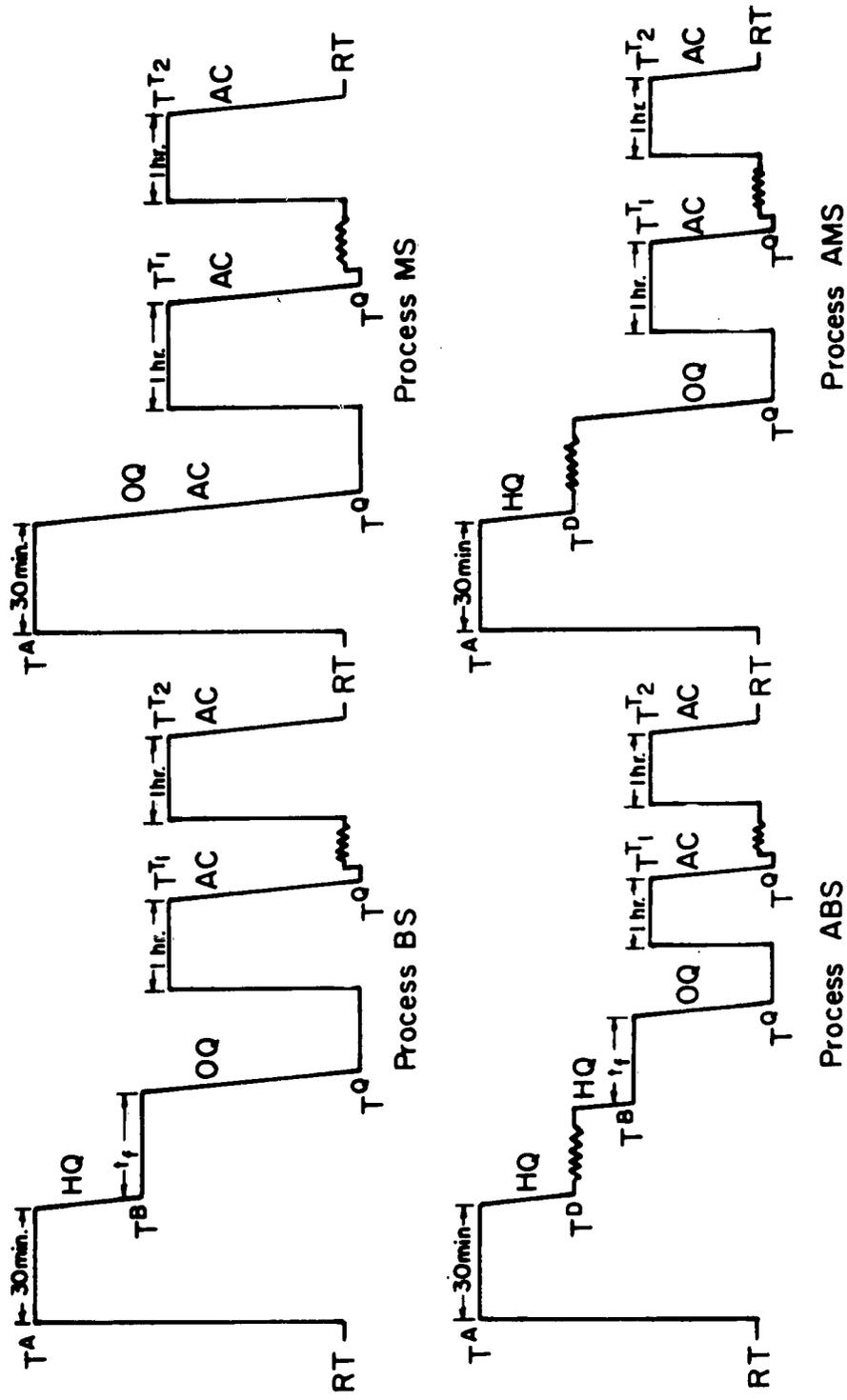


Fig. 2 - Schematic Representation of Thermomechanical Treatments.

transformation ( $t_f$ ), and oil quenching to room temperature (RT).  
Martensite was formed by air cooling H11 or oil quenching 4350 from  $T_A$ .  
Both processes B and M were followed by a double-tempering treatment  
(1 + 1 hour at  $T^T$ ) in salt baths involving air cooling to room temperature.

#### D. Deformation of Austenite

Processes AB and AM (Fig. 1) involve the deformation of austenite before transformation to either bainite or martensite. Specimens were hot quenched from  $T^A$  into a salt bath at the desired deformation temperature ( $T^D$ ). After 5 minutes in the salt, the material was rapidly transferred to an adjacent air furnace for two minutes at the same temperature ( $T^D$ ) in order to allow the salt to drain and evaporate off. Previous experience in this connection had shown that rolling a dry sample from an air furnace was easier than rolling a specimen covered with salt. The specimen was then reduced by rolling in a 2-high 15 inch mill. In the case of H11, the reduction in thickness was achieved in 4 passes, with most of the deformation occurring during the first two passes and the final two passes used primarily to straighten the blank. Because of a drop in temperature during the rolling operation, specimens were returned to the salt pot (at  $T^D$ ) for three minutes after each pass and then transferred to the air furnace (at  $T^D$ ) for two minutes. It was found that the 4350 steel had to be deformed in one pass because of the narrow time and temperature limits of the metastable austenite bay region.

The experiments involving austenite deformation were restricted to a constant reduction in thickness of 50%. Following the final deformation pass, the specimens were either hot quenched to  $T^B$  in order to form bainite or oil quenched to form martensite. Processes AB and AM were completed by double-tempering (1 + 1 hour at  $T^T$ ).

E. Strain-Tempering Martensite and Bainite

Processes BS and MS (Fig. 2) refer to straining bainite and martensite, respectively. After the formation of bainite or martensite, the structures were single-tempered (1 hour) at some temperature ( $T^{T1}$ ). At this point, deformation was performed at room temperature by rolling to the desired extent of deformation. Following the deformation the structures were retempered (1 hour at  $T^{T2}$ , where  $T^{T1}$  was not necessarily the same as  $T^{T2}$ ).

F. Combinations of Thermomechanical Treatments

Processes ABS and AMS (Fig. 2) combine deformation of the parent austenite phase with straining of the subsequently formed transformation product (either bainite or martensite).

G. Refrigeration Treatments

For various treatments, it was desirable to introduce subcooling in order to minimize or determine the influence of retained austenite on the properties. For this purpose refrigeration in liquid nitrogen was employed after the quench from either  $T^A$ ,  $T^D$  or  $T^B$ , and also after the first tempering treatment. Some treatments also required variations in the subcooling temperature ( $T^Q$ ) in the range of temperatures between room temperature and liquid nitrogen.

## H. Mechanical Testing

The hardness and unnotched tensile properties of most of the bainitic and tempered martensitic structures were determined. The parameters measured were yield strength (0.2% offset), ultimate tensile strength, elongation (2-inch gage length), and reduction in area.

A precracked subsize Charpy impact test <sup>(6)</sup> was chosen for evaluating fracture toughness. This test has been found to be relatively easy to perform and evaluate. The parameter measured was the energy per unit area required to propagate an existing crack at room temperature under impact conditions. The subsize Charpy specimens used were 2.165 inches in length, 0.394 inches in width, and 0.080 inches in thickness. A standard Charpy V-notch was ground in each specimen after heat treatment, and then a precrack (about 0.015 inch deep and extending through the specimen thickness) was produced in a fatigue apparatus specifically developed for the purpose.

## I. Metallographic Examination

Light and electron microscopy were utilized in order to correlate the observed mechanical properties with such microstructural features as prior austenite grain size, carbide and bainite morphology, and volume fraction of the various constituents present.

The times required for the initiation ( $t_g$ ) and completion ( $t_f$ ) of the bainite reaction were established by metallographic examination of isothermally reacted samples. The  $M_s$  temperature of each alloy for various austenitizing conditions was determined by the Greninger-Troiano technique. An accurate measurement of  $M_s$  was considered to be

essential since bainite reaction temperatures very close to  $M_s$  were used.

Consideration was given to quantitative determination of retained austenite by both light and electron microscopic methods. Preliminary determinations by both methods indicated that the magnitude of retained austenite content was about 25% for many treatments involving  $T^T$  of less than  $1000^\circ\text{F}$ . In a preliminary study specimens were etched with modified picral (1% HCl) solution and both light and electron micrographs were taken. Based on lineal analysis, it was found that values obtained by electron microscopy are 5 to 10% higher than by light microscopy. This is attributed to the fine state of retained austenite distribution in these specimens and the insufficient resolution obtained by light microscopy even with an oil immersion objective at 1500X.

The method finally selected for determining retained austenite contents was electron microscopy involving lineal analysis on the fluorescent screen. In order to facilitate these measurements, a device combining automatic timers with a mechanical coupler was constructed and attached to the electron microscope. This enabled the simultaneous traversing of the electron microscopic image and recording of the linear distance of each phase traversed. A linear distance corresponding to several grain diameters was traversed and counts were made of each phase. Plastic replicas for these determinations were taken from the same locations on the polished and etched specimens as was used for x-ray and light microscopy studies.

In cases where counting directly on the fluorescent screen of the electron microscope was difficult because of the very small quantity or very fine dispersion of retained austenite present, it became necessary to apply lineal analysis to electron micrographs enlarged to about 20,000X. Both the screen and micrograph measuring techniques were found to give reproducible results.

#### J. X-Ray Diffraction

Retained austenite contents were determined by the integrated intensity x-ray technique using monochromatic cr-radiation and scintillation detectors. This method is useful if the specimen is a powder or a fine-grained randomly oriented polycrystal. This method could not be employed for treatments involving austenite deformation (such as process AB) because of texturing of the austenite. However, samples from process B (not involving austenite deformation) should be susceptible to this type of analysis. Accordingly, sections were taken from the grip regions of tensile specimens subjected to process B. This area sees no plastic strain during testing and should provide a structure equivalent to untested material.

Results showed that the as-received sheet stock had considerable texture so that X-ray determinations of retained austenite could not be reliably employed even for treatments not involving austenite deformation.

### III. RESULTS AND DISCUSSION

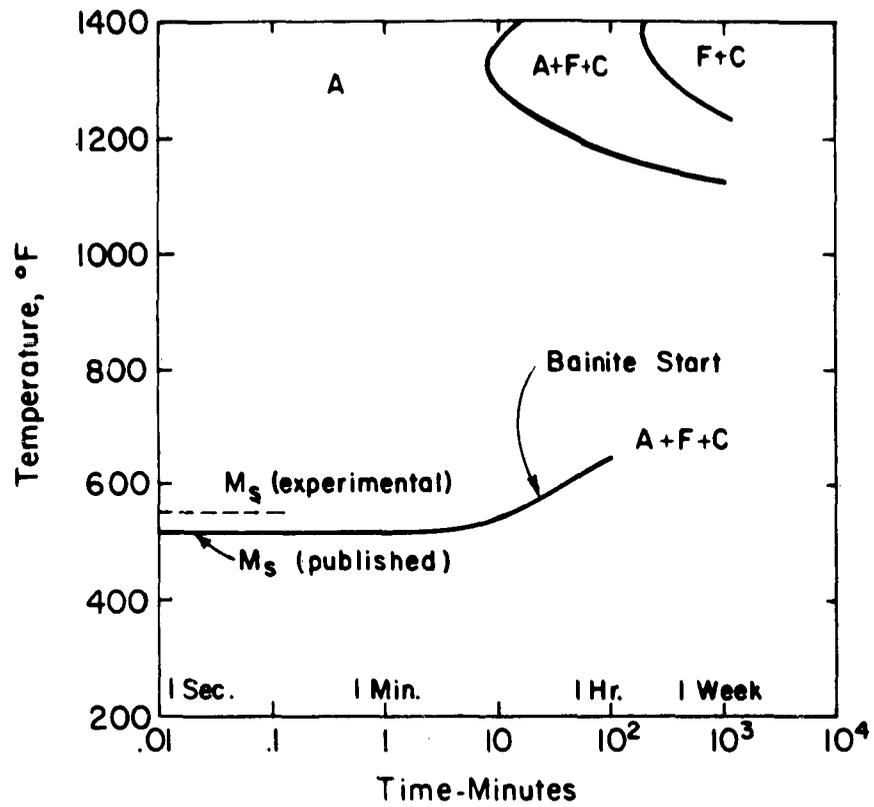
#### A. Transformation Characteristics of H11 and 4350 Steels

A study was carried out to verify and supplement the available information on the transformation characteristics of the H11 and 4350 steels. Fig. 3(a) shows a previously published isothermal-transformation diagram for H11 steel, whereas Fig. 3(b) is the result of interpolating between previously published diagrams of 4350 and 4360 steels. In line with the objectives of the present investigation, it was important to check the times to reach the end of transformation at various reaction temperatures.

For H11 steel, the bainitic transformation was studied at 50° intervals between 500 and 700°F. It was found that although no bainite forms at 700°F within 16 hours, the end of the reaction at 500, 600, and 650°F is reached in times up to 4 hours. Light and electron micrographs of specimens isothermally reacted for 4 hours at these temperatures (Figs. 4 and 5 respectively) reveal that the end of transformation at each of these temperatures does not represent 100% transformation to bainite.

The bainite morphology in the H11 steel changes and the structure coarsens as the reaction temperature is raised. In the case of the 500°F reaction, the acicular isothermal product is apparently lower bainite. Since the  $M_s$  is about 555°F, a portion of the micro structure should consist of martensite formed on cooling from 555 to 500°F; and this martensite becomes tempered during the subsequent 4-hour hold. The final micro structure also contains some untempered martensite formed on the subsequent quench from 500°F to room temperature as well as some

(a)



(b)

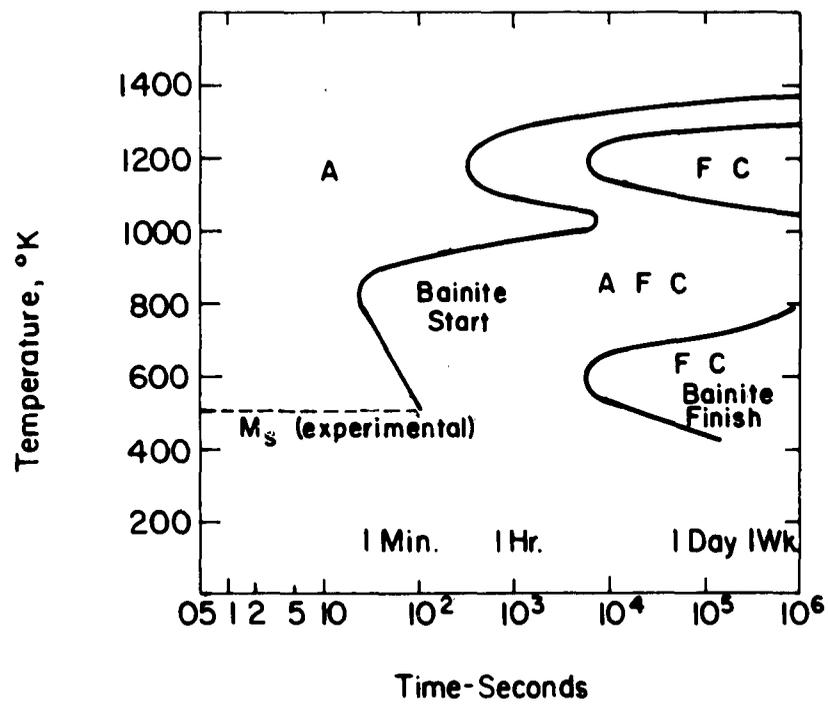
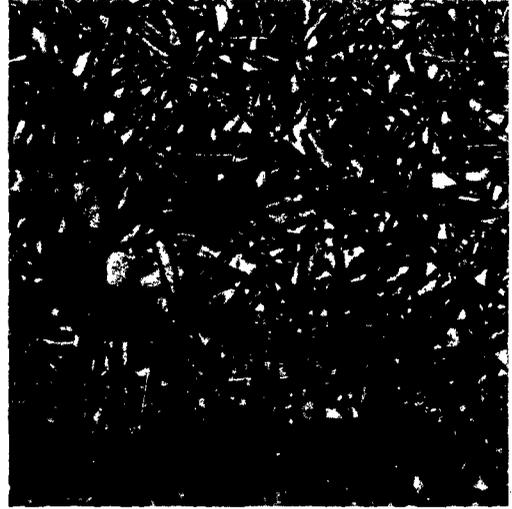


Fig. 3 - Isothermal Transformation Diagrams for (a) H11 and (b) 4350



(a)



(b)



(c)



(d)

Fig. 4 - Light Micrographs of H11 Bainite Transformed (4 hrs.) at (a) 500°F, (b) 550°F, (c) 600°F and (d) 650°F. Etchant Modified Picral. Magnification 1500X.



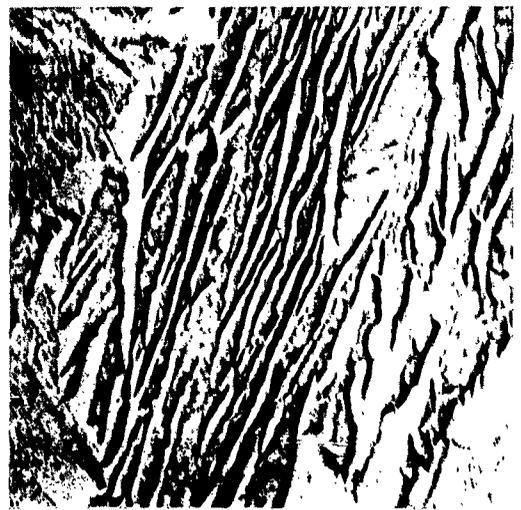
(a)



(b)



(c)



(d)

Fig. 5 - Electron Micrographs of H11 Bainite Transformed (4 hrs.) at (a) 500°F, (b) 550°F, (c) 600°F and (d) 650°F Plastic Replicas Shadowed. Etchant Modified Picral. Magnification 15,000X.

retained (un-transformed) austenite. Isothermal transformation at 550°F develops a product which cannot be definitely identified as either upper or lower bainite. Some areas in the bainite contain coarse carbides oriented approximately parallel to the long axis of the plates, as in upper bainite, while other areas contain fine carbides oriented at an angle of approximately 60 degrees to the plate axis. The remainder of the structure appears to consist of retained austenite. It is probable that small quantities of both untempered and tempered martensite are also present. The reactions at both 600 and 650°F result in upper bainite, untempered martensite and retained austenite as the final structure.

It was found that the bainitic reaction in 4350 reaches completion within 4 hours at temperatures from 550 to 800°F. The microstructure of 4350 developed by transformation at 550°F (4 hours) is illustrated in the electron micrograph of Fig. 6. In agreement with the isothermal-transformation diagram (Fig. 3b), the reaction appears to be complete, and the resulting product is lower bainite.

The martensite-start temperatures were determined by the Greninger-Troiano metallographic technique, and the results are given in Table 2.

Table 2

<u>Steel</u>	<u>M<sub>s</sub> Temperatures of the Steels</u>	
	<u>M<sub>s</sub> Temperature °F</u>	<u>Austenitizing Temperature °F</u>
H11	555	1850
4350	515	1550



Fig. 6 - Electron Micrograph of 4350 Bainite Trans-  
formed at 550°F (4 hrs.) Carbon Replica  
Shadowed. Etchant Modified Picral. Magni-  
fication 15,000X.

B. Mechanical Properties of Hll Steel Subjected to Thermomechanical Treatments

1. Bainite and Martensite - No Deformation (Processes B and M)

1.1 Bainite-Reaction Temperature ( $T^B$ )

The tensile properties of Hll as a function of bainite-reaction temperature (process B) are shown in Fig. 7. A tempering temperature ( $T^T$ ) of 550°F was employed in this series of tests. The ultimate tensile strength is found to be independent of  $T^B$  in the temperature range of 450 to 650°F and has a value of approximately 270,000 psi. However, the yield strength changes rapidly with  $T^B$  above 500°F. The decrease in the yield strength as  $T^B$  increased from 500 to 600°F is partly associated with coarsening of the bainite, as illustrated in Figs. 4 and 5. The increase in yield strength for  $T^B = 650$ °F is due to the limited amount of bainite formed and a corresponding increase in the quantity of martensite (Fig. 5d). The elongation (measured in a 2-inch gage length) increases somewhat as  $T^B$  decreases, while the reduction in area remains essentially constant.

It should be noted that specimens transformed at 450 and 500°F contain significant amounts of martensite in addition to the bainite since the  $M_s$  temperature is 550°F. This factor together with the refinement of the bainitic plate size presumably accounts for the higher yield strengths obtained at  $T^B$  below 550°F. However, in order to develop a small bainitic plate size and avoid appreciable martensite formation, a  $T^B$  of 550°F was selected for further experiments on the bainitic structures.

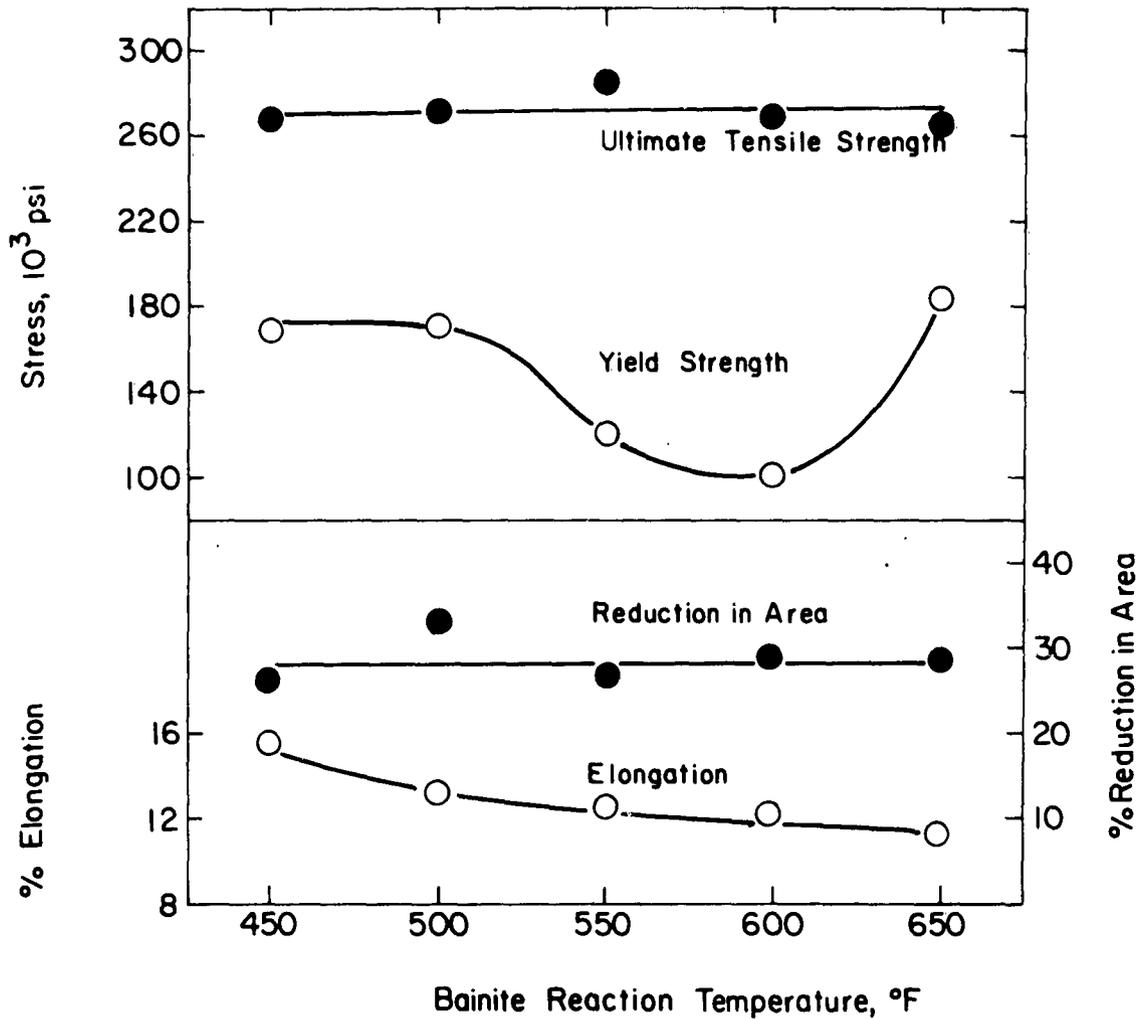


Fig. 7 - Tensile properties of H11 as a function of bainite reaction temperature ( $T^B$ ) for process B. ( $T^A=1850^\circ\text{F}$ ,  $t_f=4$  hours,  $T^Q=\text{RT}$ ,  $T^T=550^\circ\text{F}$ ).

## 1.2 Tempering Temperature ( $T^T$ )

The mechanical properties of H11 as a function of tempering temperature for process B ( $T^B = 550^\circ\text{F}$ ) are presented in Fig. 8. The effect of quench temperature ( $T^Q$ ), room temperature (RT) vs. liquid nitrogen (LN), is illustrated. The ultimate tensile strength displays a marked secondary strengthening reaction independent of  $T^Q$ , reaching a maximum value of 324,000 psi at a tempering temperature of  $1000^\circ\text{F}$ . The yield strength, for tempering temperatures of  $900^\circ\text{F}$  and below, varies between 115,000 and 150,000 psi. Tempering in the temperature range of 1000 to  $1100^\circ\text{F}$  develops a yield-strength level of 230,000 to 245,000 psi. With a LN quench an increase in yield strength of about 40,000 psi occurs for  $T^T$  below  $1000^\circ\text{F}$ . The behavior of the yield strength as the tempering temperature increases is partially associated with the transformation of retained austenite.

The ductility, as measured by tensile elongation, is unaffected by  $T^Q$  and has a value of about 12% for  $T^T$  between 550 and  $900^\circ\text{F}$ . The ductility falls to 6% for  $T^T$  values between 1000 and  $1100^\circ\text{F}$ . The reduction in area values vary between 15 and 35% with a minimum at a  $T^T$  of  $900^\circ\text{F}$ . The impact energy is low over the entire range of  $T^T$ .

Fig. 9 shows the mechanical behavior of H11 as a function of  $T^T$  for process M. Subcooling in liquid nitrogen produces significant increases in both yield and ultimate strength for  $T^T$  below approximately  $900^\circ\text{F}$ ; however, there are corresponding decreases in ductility and toughness. The effects associated with the LN treatment can be attributed to the transformation of retained austenite to martensite.

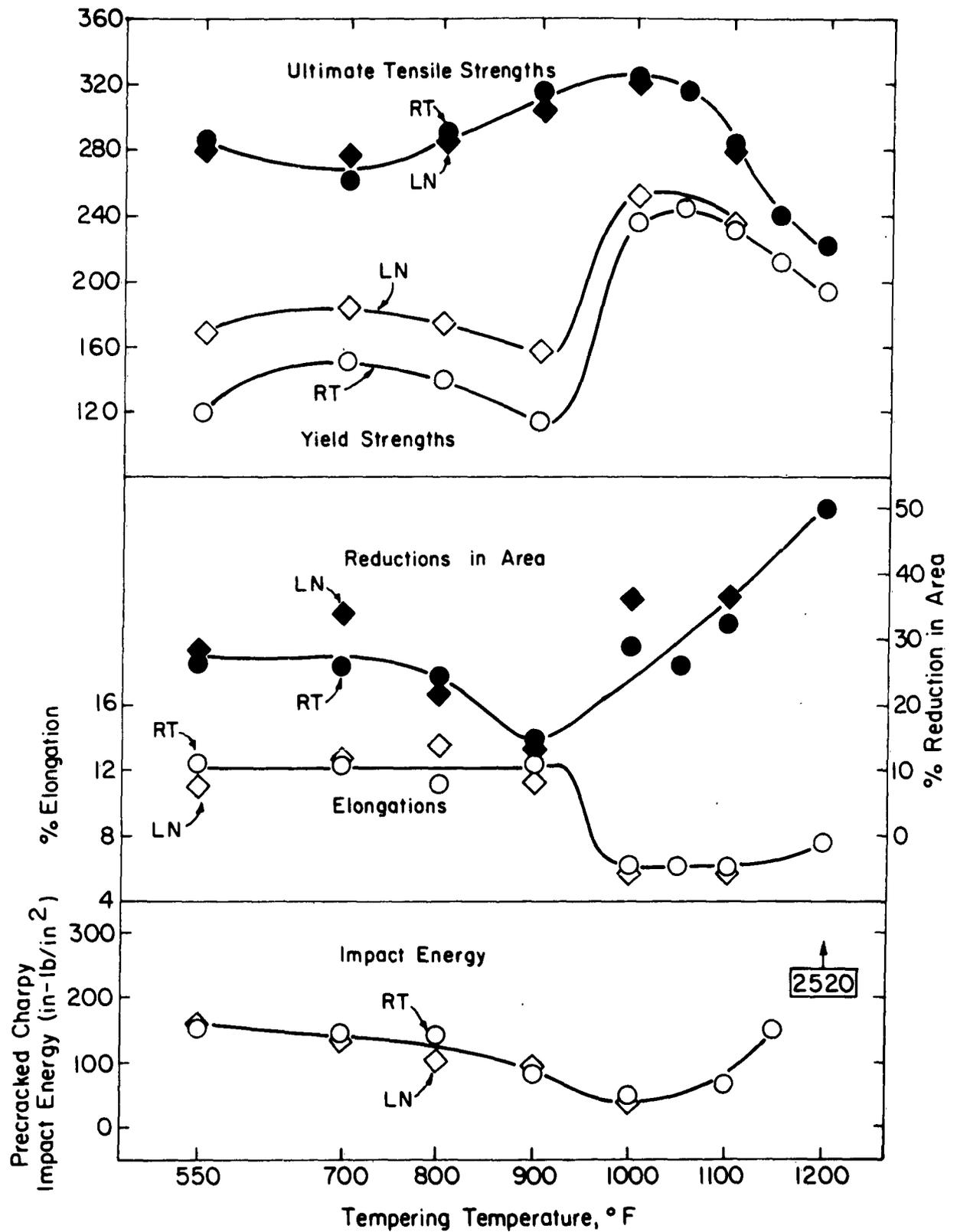


Fig. 8 - Mechanical properties of H11 as function of tempering temperature ( $T^T$  for process B,  $T^A=1850^\circ\text{F}$ ,  $T^B=550^\circ\text{F}$ ,  $t_f=4$  hours,  $T^Q=RT$  or LN).

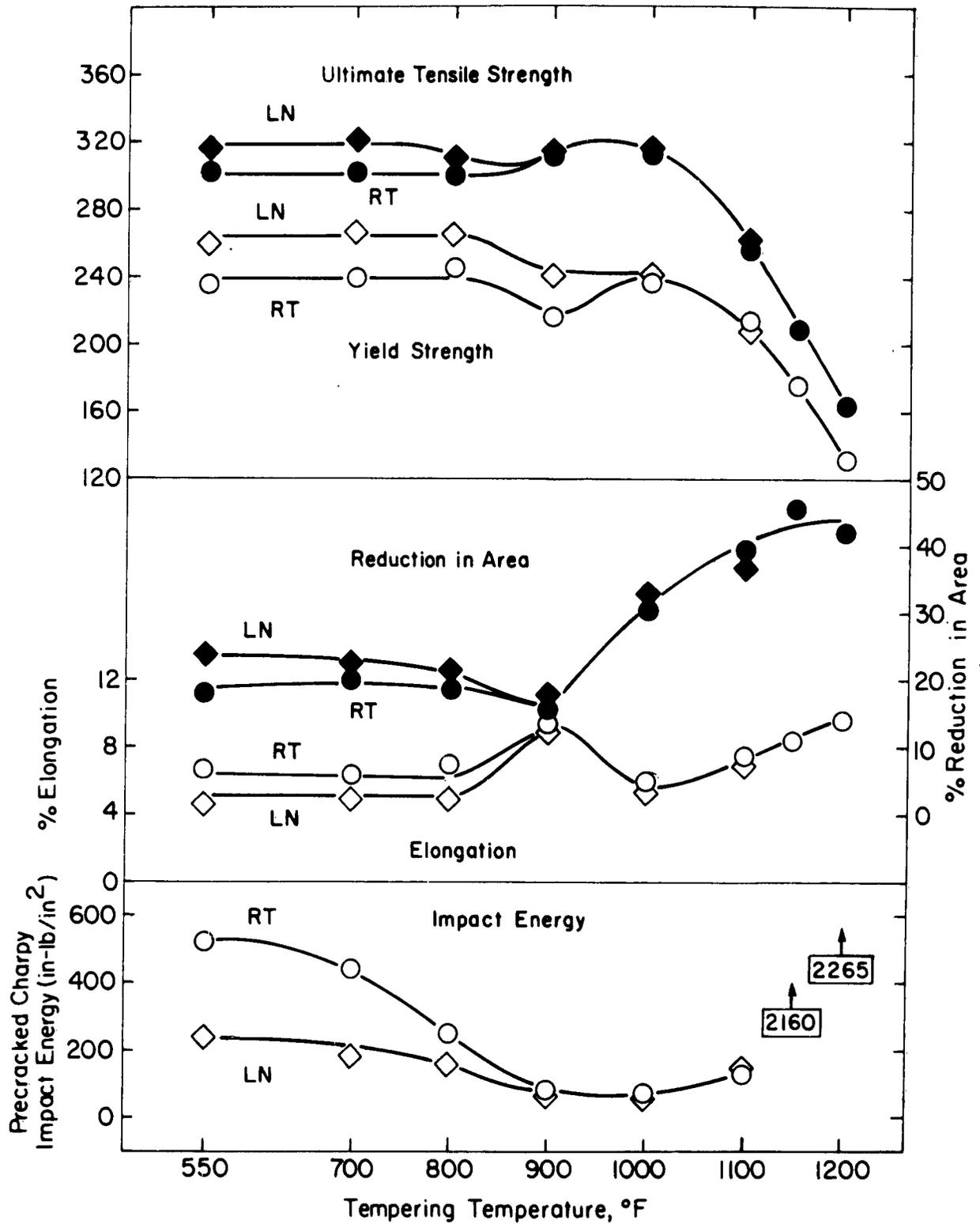


Fig.9 — Mechanical properties of H11 as a function of tempering temperature ( $T^T$ ) for process M. ( $T^A=1850^\circ\text{F}$ ,  $T^Q=\text{RT}$  or LN).

Below a  $T^T$  of  $1000^{\circ}\text{F}$ , martensite is seen to possess higher strength and toughness and lower ductility than the tempered bainites. On the other hand, for  $T^T$  above  $1000^{\circ}\text{F}$  (up to  $1200^{\circ}\text{F}$ ), the bainite has higher strength than the martensite but equivalent ductility and toughness. After an  $1100^{\circ}\text{F}$  temper the bainite has a yield strength of 236,000 psi which may be compared to martensite with a yield strength of 212,000 psi.

### 1. 3 Isothermal Transformation Time ( $t_f$ )

The mechanical properties of H11 bainite as a function of  $t_f$  at  $550^{\circ}\text{F}$  are shown in Fig. 10. The heat treatment included subsequent double tempering (1 + 1 hour) at  $550^{\circ}\text{F}$ . The effect of subcooling in liquid nitrogen is the same as previously described. The formation of bainite is accompanied by a marked loss in yield strength from the 240,000 psi level (tempered martensite) to approximately 120,000 psi, concurrent with an improvement in ductility values. This reaction ceases after about 4 hours.

Light microscopy shows that a small amount of bainite forms during the first 15 minutes at  $550^{\circ}\text{F}$  (Fig. 11). Although this structural change has no effect on the tensile properties, a significant decrease in toughness is observed. This is evidently a case where a small amount (see Fig. 11, approximately 5%) of one of the constituents in a mixed structure can play a dominant role in determining fracture toughness.

### 1. 4 Effect of Retained Austenite

The accuracy of lineal analysis applied both to electron micrographs and directly to the replicas as observed in the electron microscope was confirmed with specimens subjected to process M.

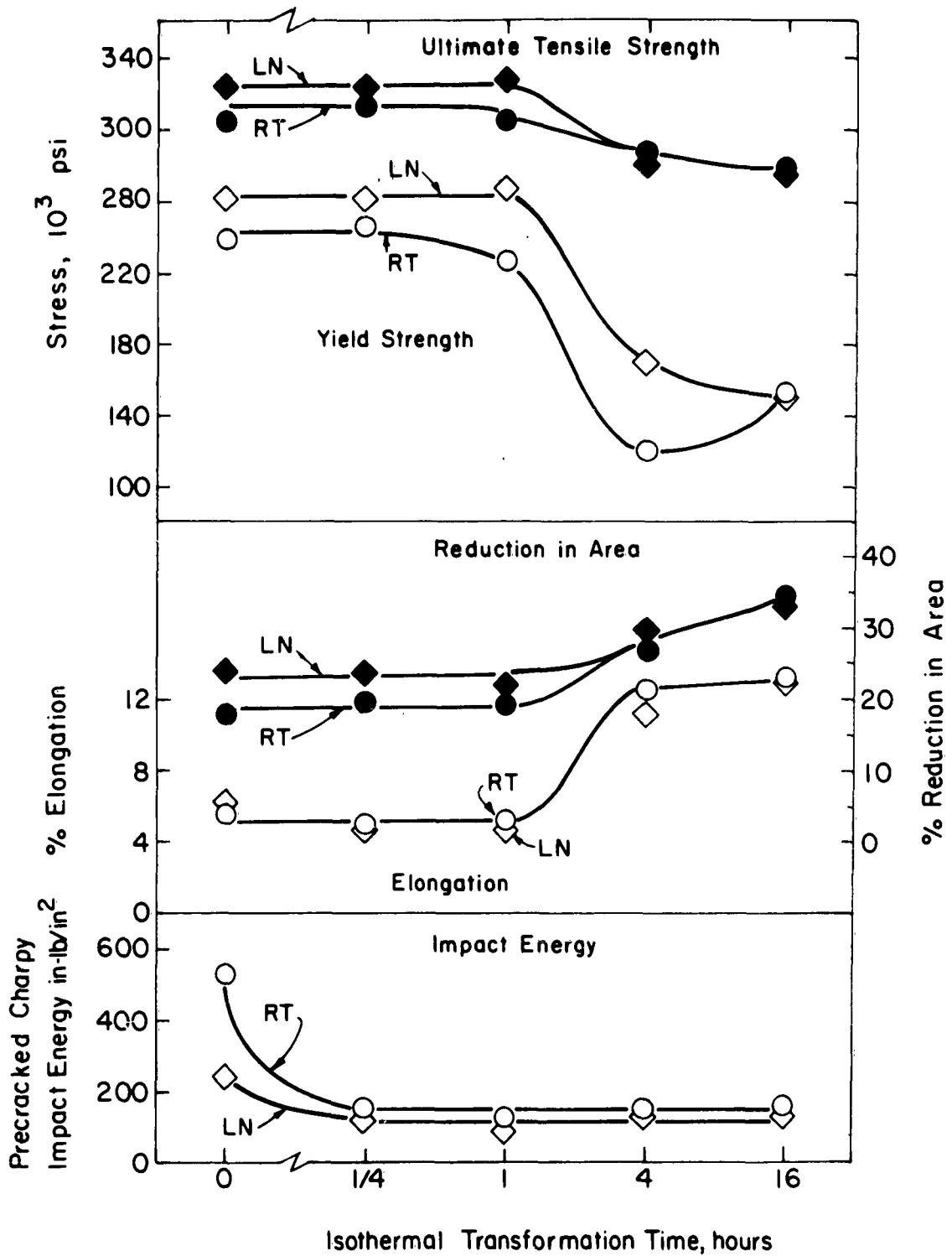


Fig. 10 - Mechanical properties of H11 as a function of Isothermal transformation time ( $t_f$ ) for process B. ( $T^A=1850^\circ\text{F}$ ,  $T^B=550^\circ\text{F}$ ,  $T^T=550^\circ\text{F}$ ,  $T^Q=$  RT or LN).



Fig. 11 - Light Micrograph of H11 Isothermally Held at 550°F for 15 Minutes and Quenched To Room Temperature. Etchant Modified Picral. Magnification 1500X.

Tempering at 600 and 1000<sup>o</sup>F gave retained-austenite values of 7.4 percent and approximately 2.0 percent respectively. These quantities agree with X-ray and magnetic determinations reported for untextured H11 similarly tempered.<sup>(11)</sup>

The details of the structure and regions of retained austenite in H11 are sharply defined for tempering temperatures below 1000<sup>o</sup>F. The electron micrographs of Fig. 12 are of H11 subjected to process B with a 550<sup>o</sup>F temper. The "smooth" austenite regions are easily distinguished from the bainite plates which were formed at 550<sup>o</sup>F.

The variation of yield strength and retained austenite content of H11 as a function of tempering temperature for process B is shown in Fig. 12. Tempering in the range 550 to 900<sup>o</sup>F with T<sup>Q</sup> at RT develops yield strengths of 120,000 to 155,000 psi. The presence of approximately 35 percent retained austenite in the bainitic structures tempered in this range probably accounts for the low strengths. This austenite may be partially converted to martensite by subcooling in liquid nitrogen. The refrigeration lowers the austenite content to 25 percent and raises the yield strengths to the range of 160,000 to 180,000 psi. Increasing the tempering temperature to 1000<sup>o</sup>F causes a large increase in yield strength concurrent with a decrease in austenite content for both unrefrigerated and refrigerated specimens.

It should be noted that the ultimate tensile strength is not affected by refrigeration treatments over this tempering range (Fig. 8). In addition, there is a large difference between yield and ultimate strengths.



(a)



(b)

Fig. 12 - Electron Micrographs of H11 for Process B  
Tempered at 550°F. Plastic Replicas, Shadowed  
Etchant Modified Picral. Magnification (a) 6,700X,  
(b) 15,000X.

These effects may be associated with conversion of some of the austenite during the tensile test. The degree of austenite conversion between yielding and the onset of necking is a problem that requires further investigation.

The data presented in Fig. 13 may be considered in terms of the increments in yield strength and austenite content as a function of tempering temperature. Figure 14 shows that the increase of yield strength is approximately proportional to the decrease in percent austenite for tempering temperatures up to 900<sup>o</sup>F. However, at 1000<sup>o</sup>F there is a large change in austenite content upon subcooling but only a small improvement in yield strength. This apparent discrepancy may be explained if the stability of the austenite after tempering at 1000<sup>o</sup>F is such that it is converted by very low strains. If strains below 0.2 percent can convert a large portion of the original 16 percent austenite, then the yield strengths of refrigerated and unrefrigerated samples would be similar. The fact that more austenite transforms upon subcooling with a 1000<sup>o</sup>F temper than with lower tempers is an indication that the austenite retained after this temper without refrigeration is less stable than that obtained after lower tempers.

It was pointed out in connection with Figs. 8 and 9 that process B structures are as strong or stronger than process M structures after tempering at 1000 to 1200<sup>o</sup>F, but not below 1000<sup>o</sup>F. The high austenite contents of bainitic structures tempered below 1000<sup>o</sup>F accounts for the lower strengths. For example, tempering at 550<sup>o</sup>F with process M develops a structure with 7 percent austenite having a yield strength of

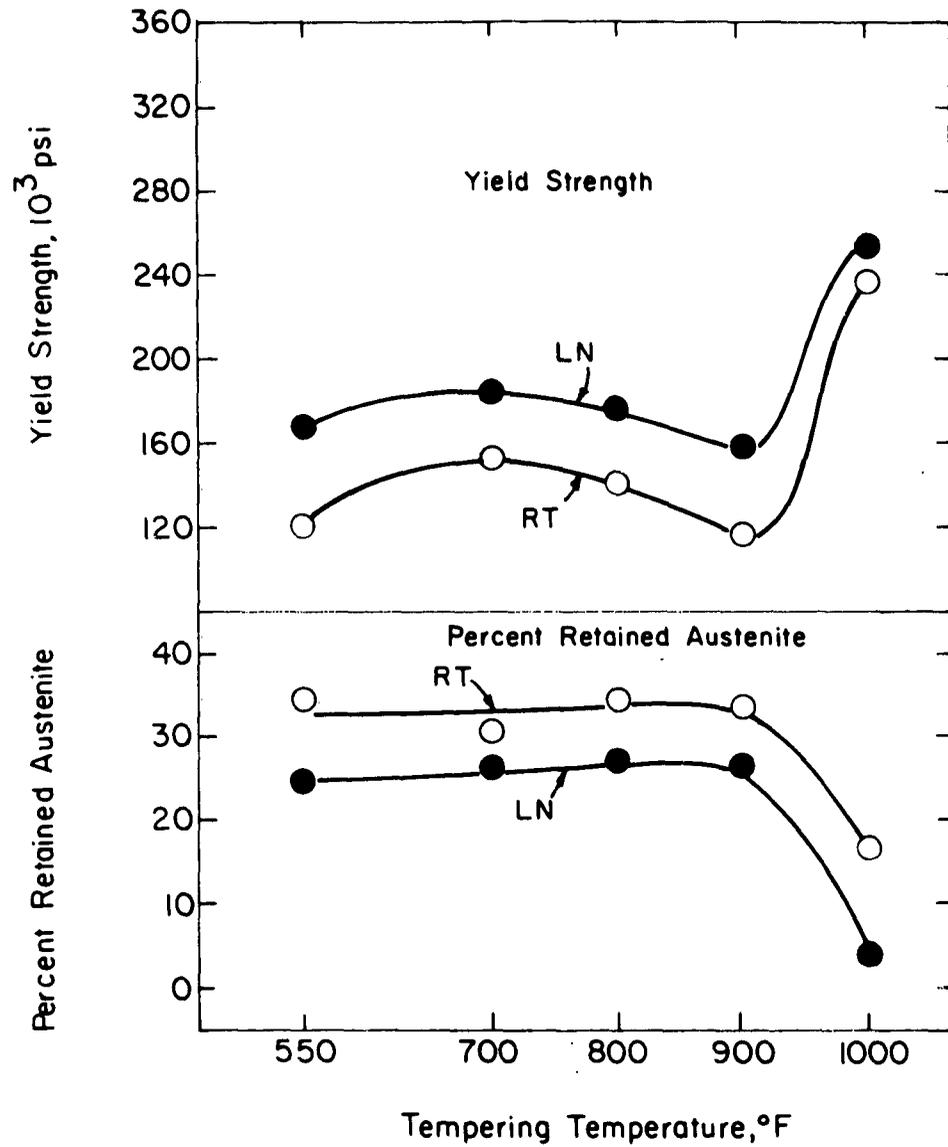


Fig.13-Yield strength and retained austenite content of H11 as a function of tempering temperature for process B. ( $T^A=1850^\circ\text{F}$   $T^B=550^\circ\text{F}$ ,  $t=4$  hours,  $T^Q=RT$  or LN).

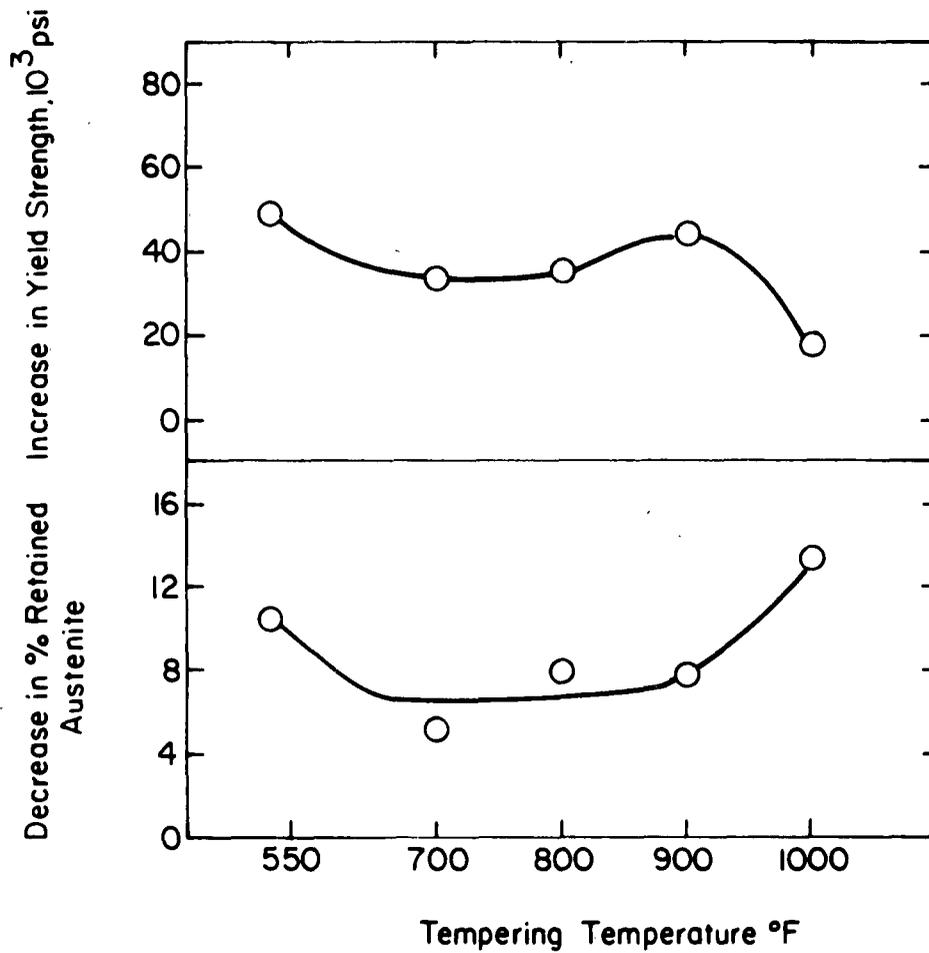


Fig.14- Effect of refrigeration treatment (LN) on yield strength and retained austenite content of H II subjected to process B. ( $T^A=1850^\circ\text{F}$ ,  $T^B=550^\circ\text{F}$ ,  $t = 4$  hours,  $T^Q=RT$  or LN).

238,000 psi as compared to 34 percent retained austenite in process B which has a yield strength of 119,000 psi. Tempering at 1000°F followed by refrigeration produces essentially the same strengths (240,000 psi) and similar retained-austenite contents for both processes M and B. One can therefore conclude that the bainite is at least as strong as the martensite under these conditions.

The influence of retained austenite on the yield strength cannot be directly proportional to its volume percent. The data plotted in Fig. 15 assumes that the strength of the bainite-martensite-austenite structure is directly proportional to the volume percent of austenite or that the strengthening due to austenite conversion is a linear function of the volume converted. Consider the data for the 550°F tempered. The highest strength that could be attached to the non-austenite part of a mixed structure quenched to room temperature is 180,000 psi. This assumes the austenite has zero strength. If this were valid, the martensite formed on subcooling would have a yield strength of at least 480,000 psi as determined from the slopes of the lines. It should be noted that the structure consisting of zero austenite is not 100 percent bainite but a mixture of bainite and martensite. Therefore, these data demonstrate that the size and distribution of the retained austenite regions are of primary importance. Furthermore, there may be a range of austenite content where the influence on strength of the mixed structures is most critical. This could provide another explanation for the apparent discrepancy in the results on tempering at 1000°F. Finally, the different slopes in Fig. 15 suggest that tempering temperature is an important variable in the influence of austenite with respect to the strength of mixed structures.

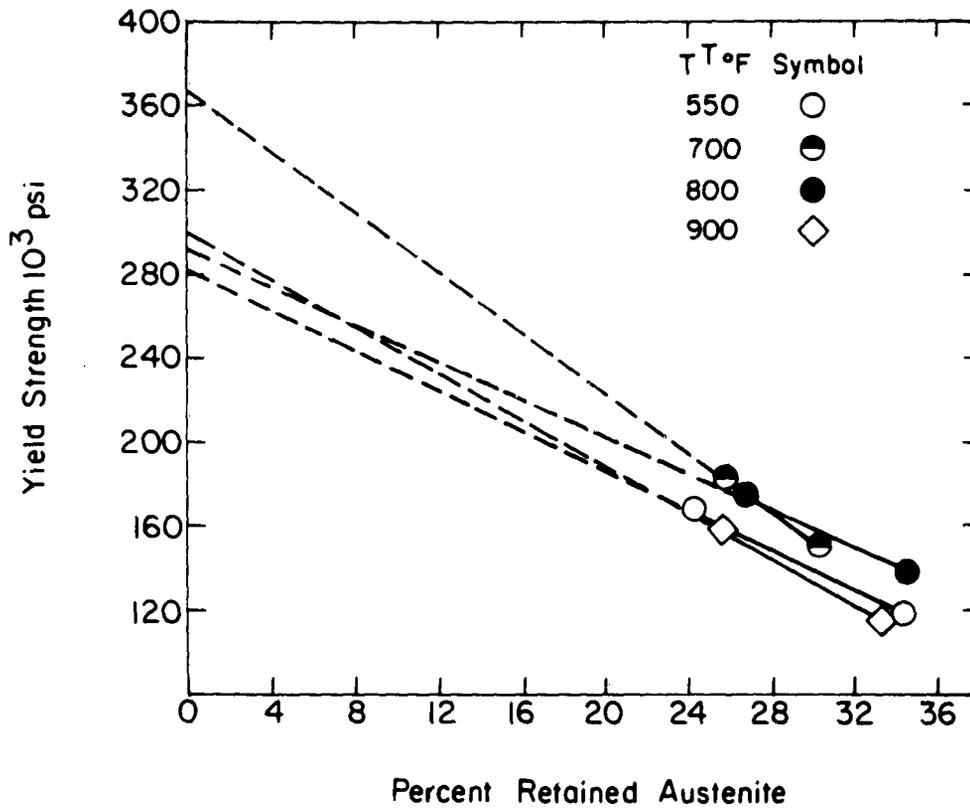


Fig.15- Yield strength H11 as a function of percent retained austenite for various tempering temperatures in process B. ( $T^A=1850^\circ\text{F}$ ,  $T^B=550^\circ\text{F}$ ,  $t_f=4$  hours,  $T^Q=\text{RT or LN}$ ).

## 2. Austenite Deformation Prior to Transformation (Processes AB and AM)

### 2. 1 Bainite-Reaction Temperature ( $T^B$ )

In order to study processes involving austenite deformation and subsequent bainite formation in the vicinity of  $M_s$ , it is important to know the effect of deformation on the progress of the martensitic reaction. Fig. 16 shows the extent of martensite formation from both deformed and undeformed austenite as a function of quench temperature. Although 50% deformation of austenite was found to depress the  $M_s$  only slightly, the shape of the subsequent transformation curve is significantly altered.

Fig. 17 depicts the tensile properties of H11 as a function of the bainite-reaction temperature for process AB. The LN treatments markedly improve the yield strength for all bainite-reaction temperatures. Variation of  $T^B$  and  $T^Q$  produces a number of interesting combinations of strength and ductility. For example, a treatment consisting of a)  $T^B = 550^\circ\text{F}$ , b) subcooling in liquid nitrogen, and c) subsequent tempering at  $550^\circ\text{F}$  develops a yield strength of 250,000 psi, an ultimate strength of 350,000, an elongation of 11%, and a RA of 27%. This may be compared with the properties of conventionally heat treated H11, i. e. yield strength of 240,000 psi, ultimate strength of 315,000 psi, and an elongation of about 5-1/2%. Thus, a  $T^B$  of  $550^\circ\text{F}$  was used for further experiments involving bainite formation from deformed austenite.

### 2. 2 Tempering Temperature ( $T^T$ )

Fig. 18 shows the mechanical properties of H11 as a function of tempering temperature for process AB where  $T^B$  is  $550^\circ\text{F}$ .

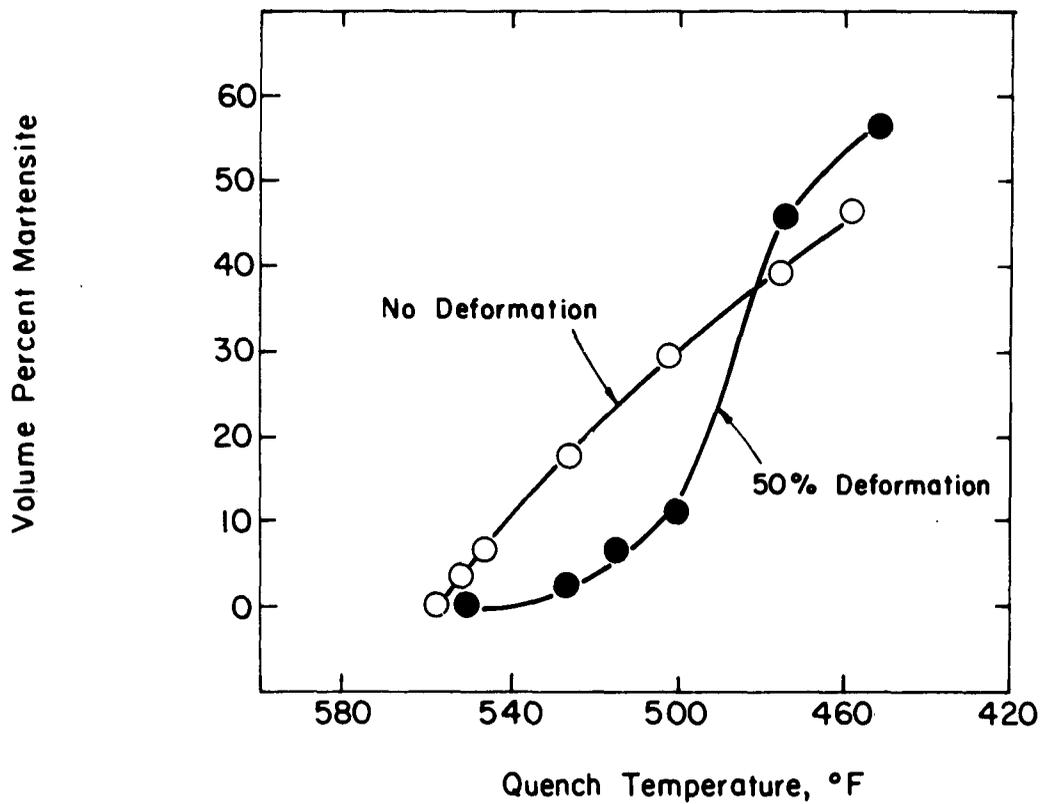


Fig. 16- Progress of martensitic reaction in H11 for deformed and undeformed austenite. ( $T^A=1850^\circ\text{F}$ ,  $T^D=900^\circ\text{F}$ , deformation = 50%).

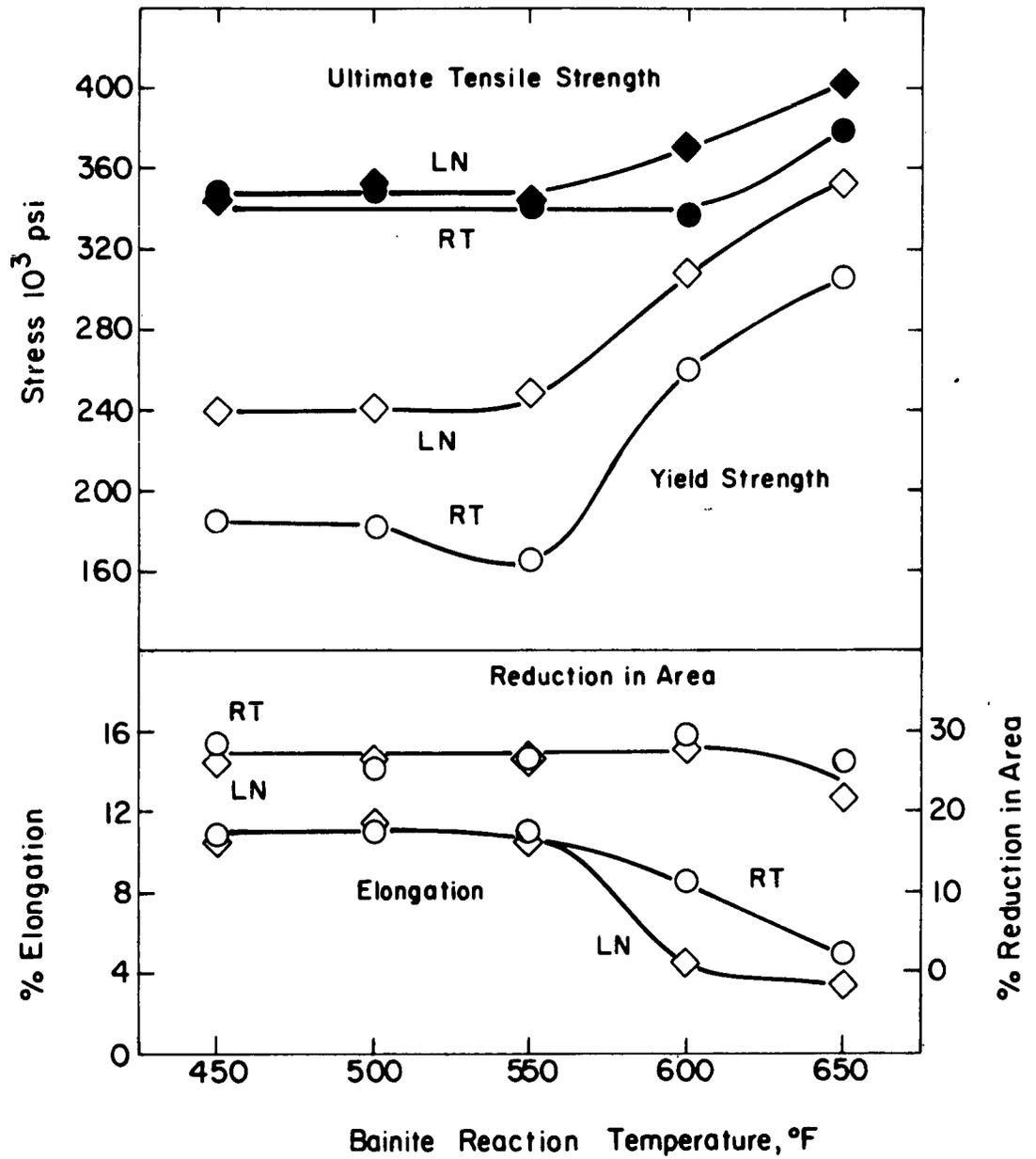


Fig. 17 - Tensile properties of H11 as a function bainite reaction temperature ( $T^B$ ) for process AB. ( $T^A=1850^\circ\text{F}$ ,  $T^D=900^\circ\text{F}$ , deformation = 50%,  $t_f = 4$  hours  $T^Q=RT$  or LN,  $T^T=550^\circ\text{F}$ ).

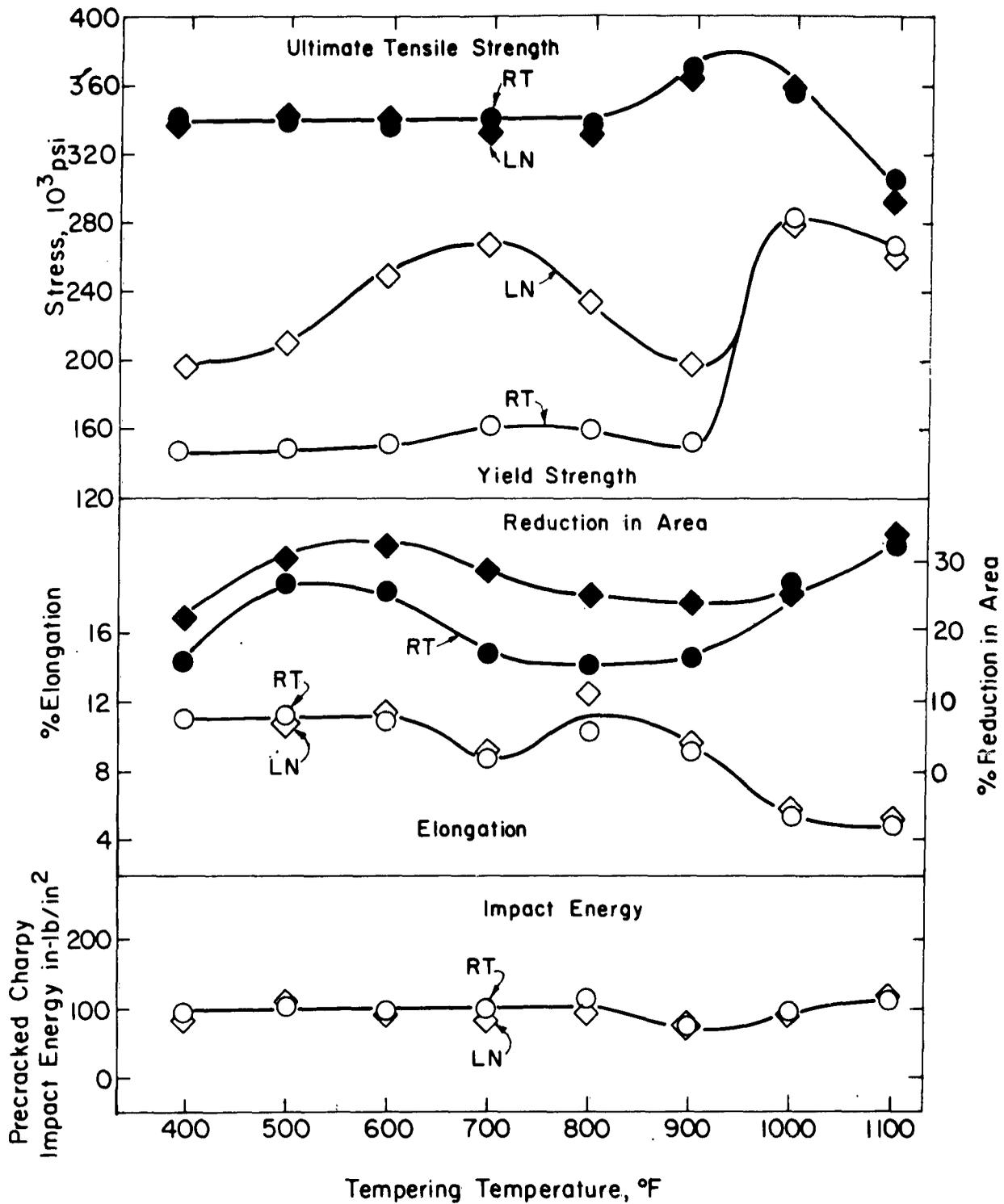


Fig.18-Mechanical properties of H11 as a function of tempering temperature ( $T^T$ ) for process AB. ( $T^A=1850^\circ\text{F}$ ,  $T^D=900^\circ\text{F}$ , deformation =50%,  $T^B=550^\circ\text{F}$ ,  $t=4$  hours,  $T^Q=RT$  or LN).

Although the trends in mechanical properties are rather similar to those observed without austenite deformation (Fig. 8), much greater strength is obtained with deformation. Also, the effect of LN treatments on the yield strength is more pronounced in this case. With a LN quench, yield strengths over 200,000 psi, ultimate tensile strengths of about 340,000 psi, and elongation values between 9 and 11% are obtained for all values of  $T^T$  up to 900°F. Tempering above 900°F produces marked secondary strengthening, which may be associated with the conditioning and subsequent elimination of the retained austenite. The impact-energy values are low and seem insensitive to both  $T^T$  and  $T^Q$ .

The effect of  $T^T$  on the tensile properties of H11 martensite produced by process AM is shown in Fig. 19. These ausformed properties display higher strength and lower ductility than was obtained by process AB (Fig. 14) for  $T^T$  less than 1000°F. However, at the higher values of  $T^T$ , the bainitic structures exhibit superior strength. For example, on tempering at 1100°F, process AB develops a yield strength of 260,000 psi and an ultimate tensile strength of 300,000 psi as compared to a yield strength of 240,000 psi and an ultimate tensile strength of 275,000 psi obtained by ausforming. Both processes produce elongations of about 5% for an 1100°F temper.

When the quench temperature ( $T^Q$ ) is varied, a change in properties is obtained as shown in Fig. 20, (for  $T^B = 550^\circ\text{F}$  and  $T^T = 550^\circ\text{F}$ ). The yield strength increases continuously with decreasing  $T^Q$  between RT and LN. This suggests that the retained austenite is relatively unstable in the vicinity of room temperature, and undergoes

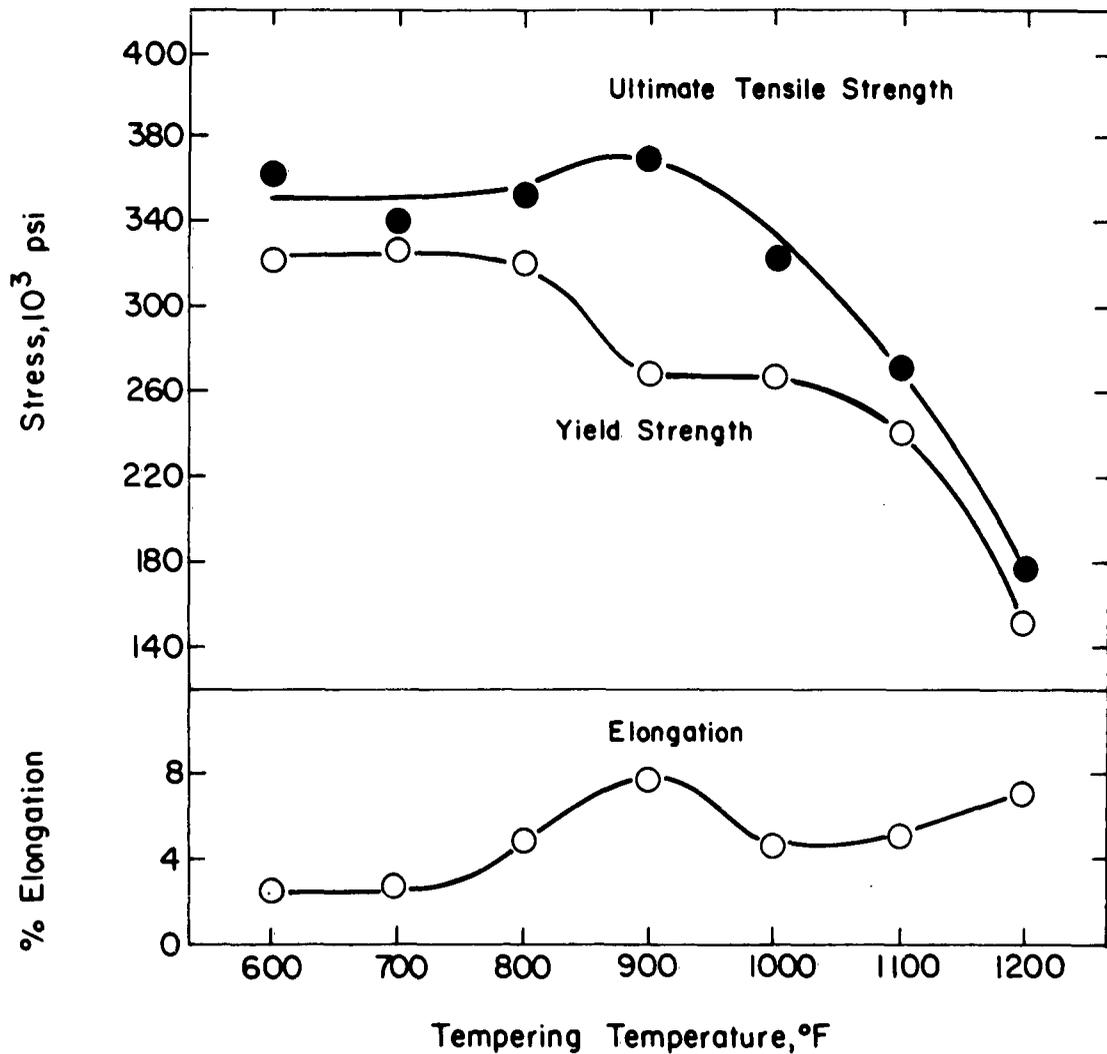


Fig.19 - Tensile properties of H11 as a function of tempering temperature, ( $T^T$ ) for process AM. ( $T^A=1850^\circ\text{F}$ ,  $T^D=1000^\circ\text{F}$ , deformation = 50%,  $T^Q=\text{RT}$ ). Courtesy of Pratt and Whitney Aircraft, United Aircraft Corporation.

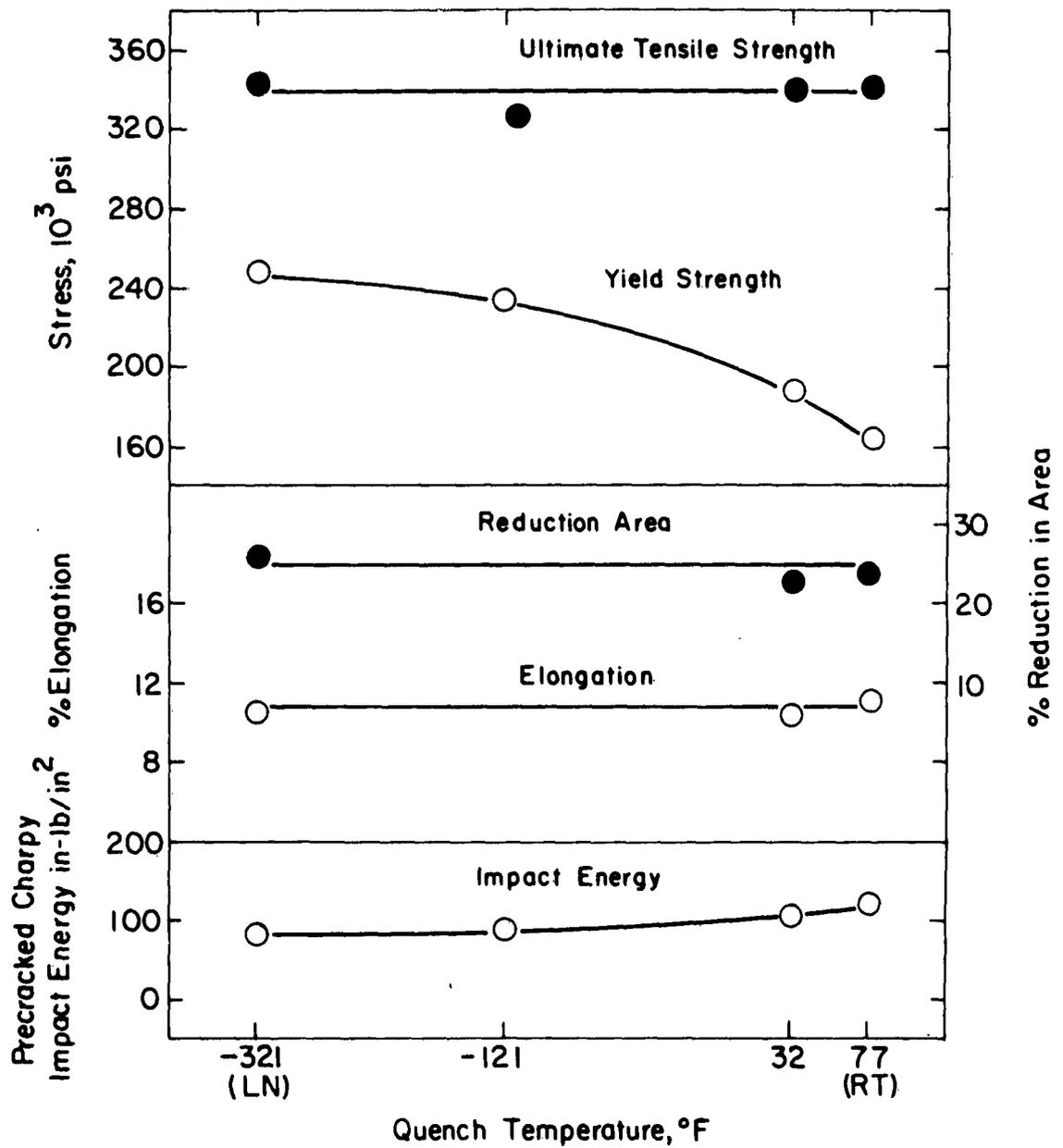


Fig. 20 - Mechanical properties of H II as a function of quench temperature ( $T^Q$ ) for process AB. ( $T^A=1850^\circ\text{F}$ ,  $T^D=900^\circ\text{F}$ , deformation = 50%,  $T^B=550^\circ\text{F}$ ,  $t_f=4$  hours  $T^T=550^\circ\text{F}$ ).

some conversion to martensite during room-temperature tensile testing of process AB specimens. This situation could account for the relatively large difference between the yield and ultimate strengths of the bainitic structures.

### 2. 3 Isothermal Transformation Time ( $t_f$ )

The effect of bainite-reaction time at 550°F in process AB on the mechanical properties of H11 are depicted in Fig. 21. The parameters most sensitive to both  $t_f$  and  $T^Q$  are the yield strength and elongation. With  $T^Q$  at RT, the yield strength decreases and the elongation increases when  $t_f$  is greater than about 15 minutes. However, when  $T^Q$  is at LN, the yield strength and elongation do not begin to change until  $t_f$  exceeds 1 hour.

The results shown in Fig. 21 together with the tempering series (Fig. 14) indicate that the formation of bainite in H11 is accompanied by the retention of austenite. Evidently, this austenite is a controlling factor in determining the yield strength and elongation characteristics. It is difficult to evaluate the inherent properties of the bainite unless essentially all the retained austenite is eliminated.

### 2. 4 Effect of Retained Austenite

The influence of austenite deformation on the resulting bainitic structure for process AB is shown in the electron micrographs of Fig. 22 (compare with Fig. 12). The bainitic plates (formed at 550°F) and austenitic regions are reduced in size compared to that obtained in process B, and the austenite is not as easily distinguished at the low magnifications, (Fig. 12a and Fig. 22a). However, at the higher

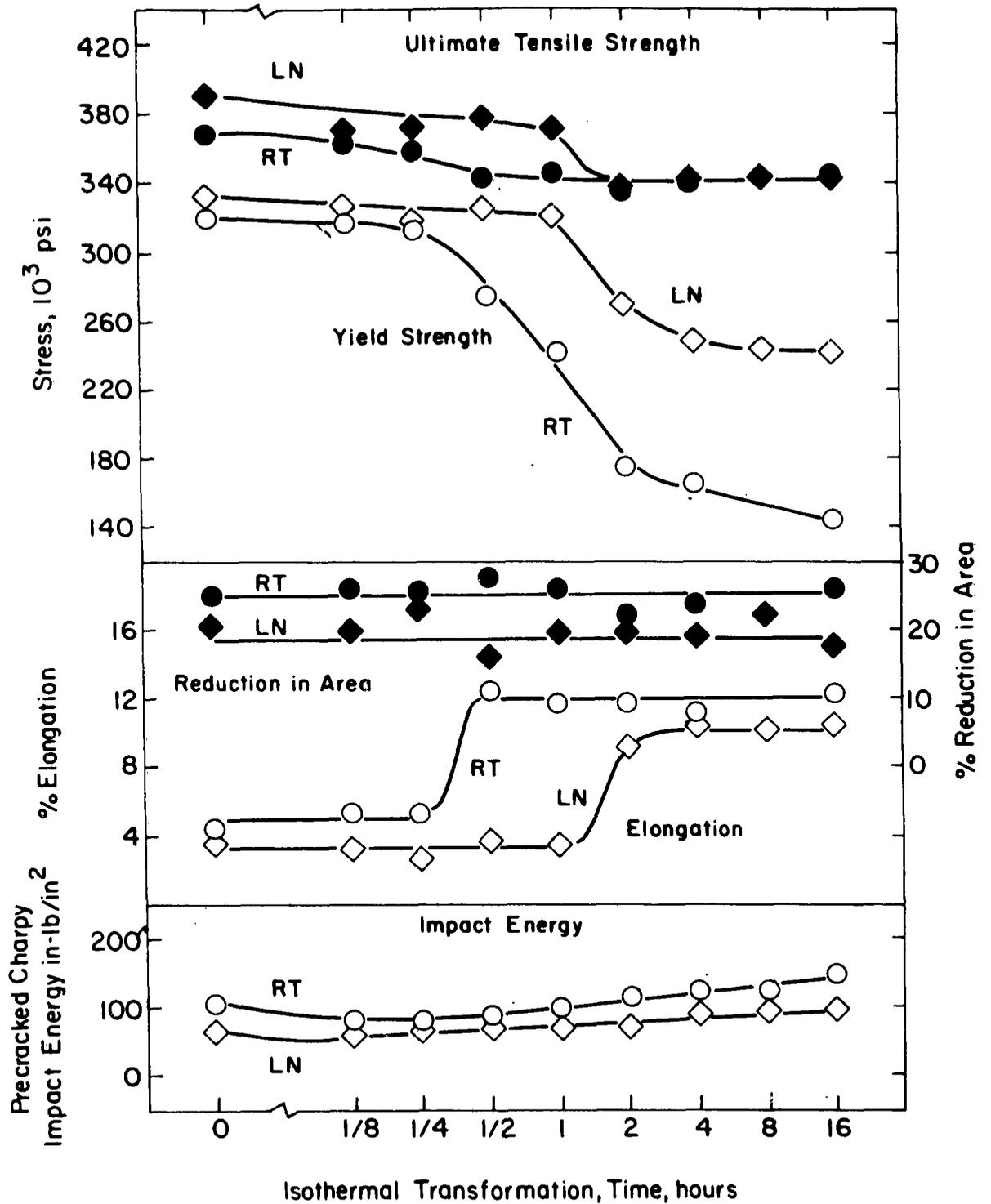


Fig. 21 - Mechanical properties of H11 as a function of isothermal transformation time ( $t_f$ ) for process AB. ( $T^A=1850^\circ\text{F}$ ,  $T^D=900^\circ\text{F}$ , deformation = 50%,  $T^B=550^\circ\text{F}$ ,  $T^Q=\text{RT or LN}$ ,  $T^T=550^\circ\text{F}$ ).



(a)



(b)

Fig. 22 - Electron Micrographs of H11 for Process AB  
Tempered at 550°F. Plastic Replicas, Shadowed.  
Etchant Modified Picral. Magnification (a) 6,700X,  
(b) 15,000X.

magnification (Fig. 22b), the austenite is well defined and an accurate determination of the volume fraction may be made.

The yield strength and retained austenite of H11 as a function of tempering temperature for process AB are given in Fig. 23. Without refrigeration, the amounts of retained austenite are comparable to those produced with process B for tempering temperatures below 1000<sup>o</sup>F. However, over this tempering range, the yield strengths are higher for process AB, illustrating the beneficial effect of prior austenite deformation on the bainite strength. Applying a subcooling treatment to process AB converts more austenite to martensite than in process B at corresponding tempering temperatures. In addition, the martensite that forms upon refrigeration from the deformed austenite should have the higher yield strength typical of "ausformed" (process AM) martensite. These two considerations make refrigeration treatments for process AB more effective in improving the yield strength than for process B. The increase in yield strength with subcooling appears more dependent upon tempering temperature in process AB than in process B. With a 700<sup>o</sup>F temper these factors produce the maximum increase in yield strength when a refrigeration treatment is combined with process AB. As shown in Fig. 24 for a 700<sup>o</sup>F temper, an increment of 105,000 psi and a decrease of approximately 14 percent in retained austenite content are developed by refrigeration in liquid nitrogen. Tempering at 1000<sup>o</sup>F causes a marked increase in yield strength and decrease in retained austenite content from that observed with a 900<sup>o</sup>F temper. In contrast to the 1000<sup>o</sup>F temper for process B, there is no appreciable difference between the yield strengths of refrigerated and unrefrigerated material for process AB (1000<sup>o</sup>F temper).

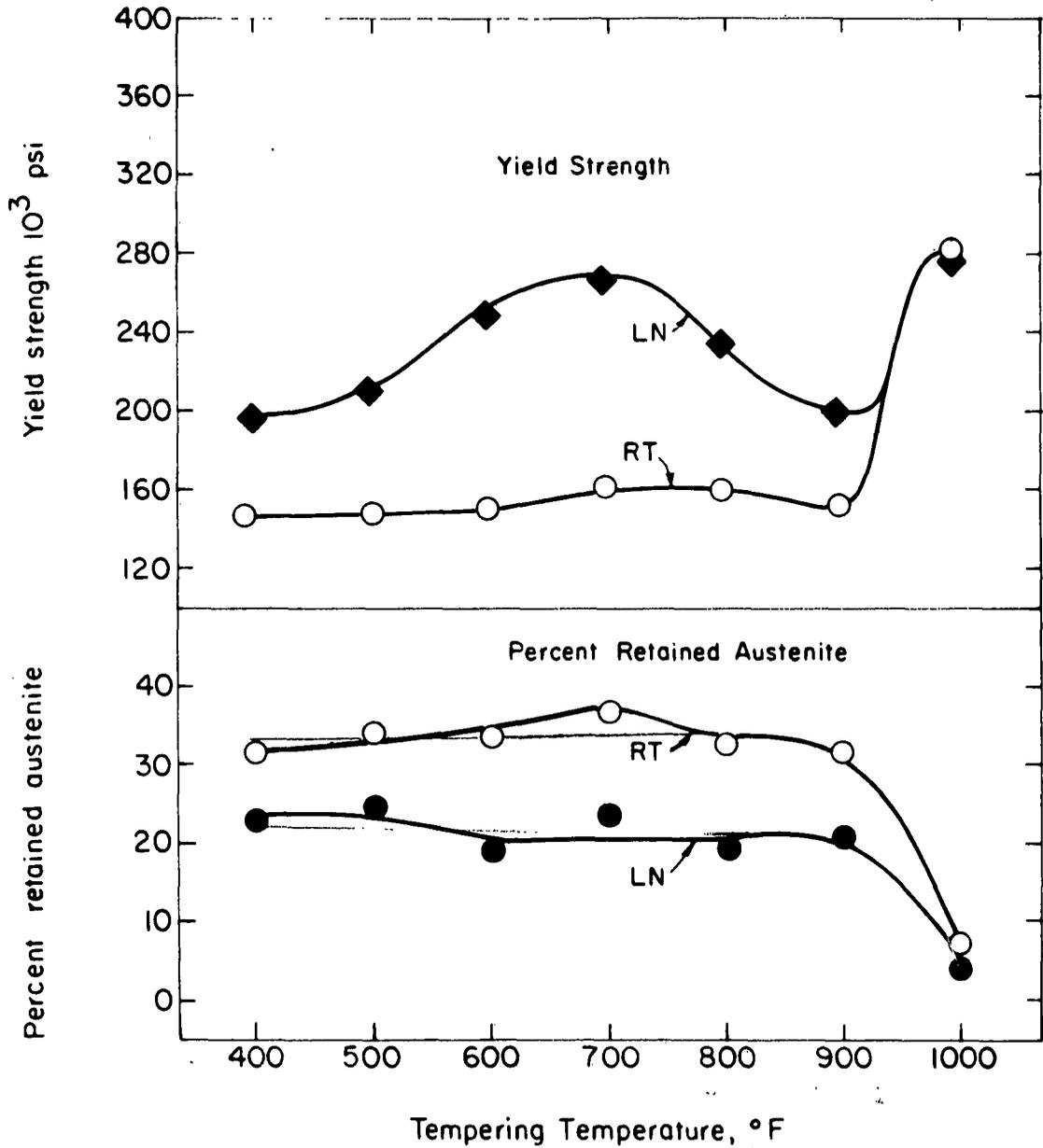


Fig.23- Yield strength and retained austenite content of H II as function of tempering temperature for process AB. ( $T^A=1850^\circ\text{F}$ ,  $T^D=900^\circ\text{F}$ , deformation 50%,  $T^Q=550^\circ\text{F}$ ,  $t_f = 4$  hours,  $T = \text{RT or LN}$ ).

Over the entire tempering range in Fig. 24, the curve for the increment in yield strength is rather parallel to the curve for the decrease in retained austenite. Again, if it is assumed that the yield strength of H11 is a linear function of the percent retained austenite for each tempering temperature, as shown in Fig. 25, one obtains a wide range of values for the strength of a zero austenite structure. Furthermore the slopes for most of the temperatures are such that the transformed product would have a yield strength at least 600,000 psi. Therefore, as in the case of process B, the influence of retained austenite in weakening the bainitic structure is not directly proportional to the amount of austenite. Similarly, the indications are that the size and distribution of the austenite are of primary importance in its weakening effect on the bainitic structure.

### 3. Deformation of Bainite (Process BS)

#### 3. 1 Percent Deformation

The mechanical properties of H11 as a function of percent reduction in thickness for process BS are presented in Fig. 26. The bainite was formed at 550°F, pretempered ( $T^{T1}$ ) at 550°F, strained by rolling at room temperature, and retempered ( $T^{T2}$ ) at 550°F. Deformation of only 3.5% followed by retempering raises the yield strength from 120,000 to 275,000 psi, and decreases the elongation from 12 to 5%. There is also a small improvement in the ultimate tensile strength from 275,000 psi to 285,000 psi. Further deformation beyond 3.5% increases the yield strength slightly faster than the ultimate tensile strength, so that the yield-to-ultimate ratio approaches unity. A deformation of 50% develops a yield strength of 370,000 psi, an ultimate tensile strength of

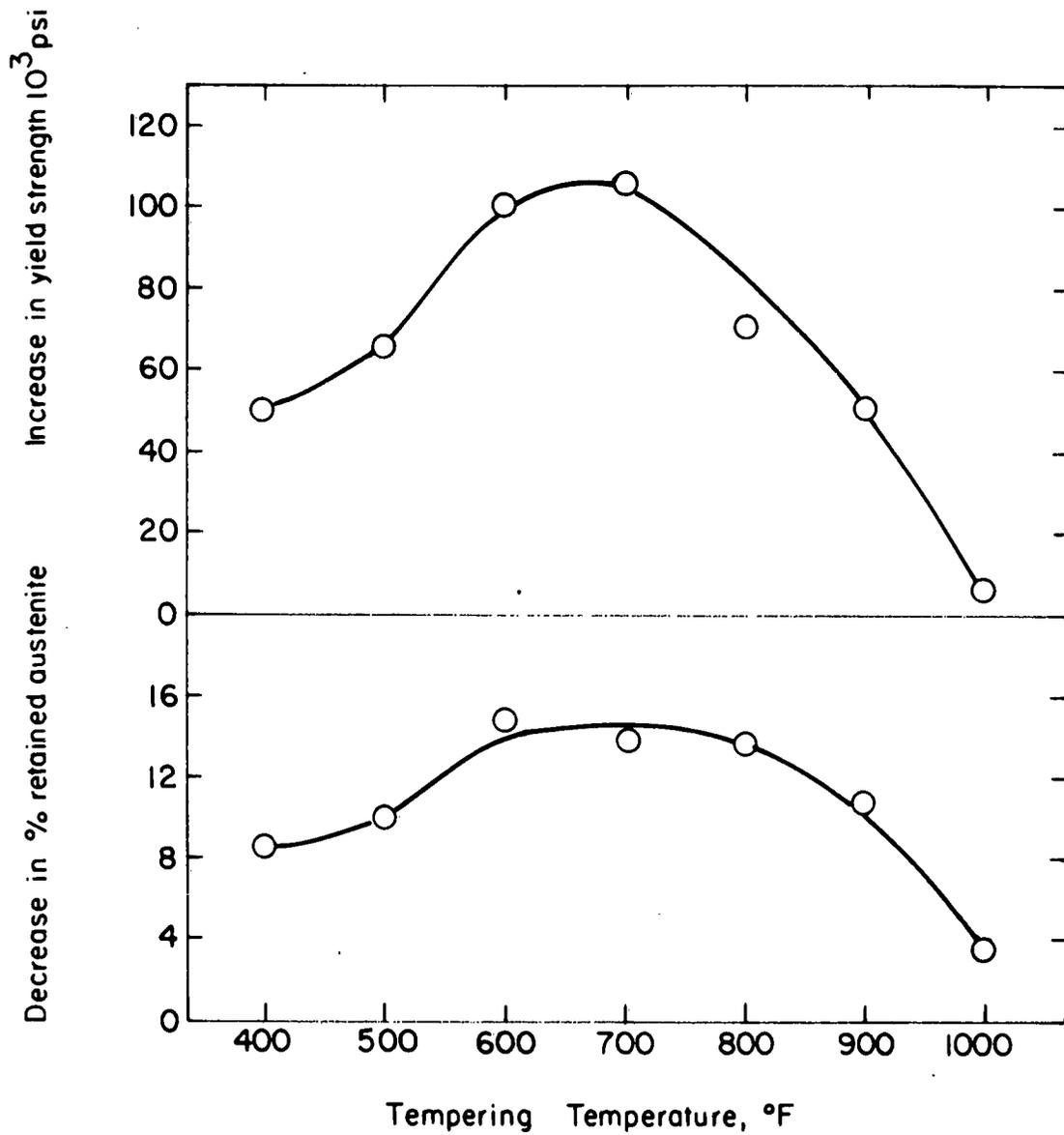


Fig.24-Effect of refrigeration treatment (LN) on yield strength and retained austenite content of H11 subjected to process AB. ( $T^A=1850^\circ\text{F}$   $T^B=900^\circ\text{F}$  deformation 50%  $T^B=550^\circ\text{F}$   $t_f=4$  hours  $T^Q=\text{RT or LN}$ ).

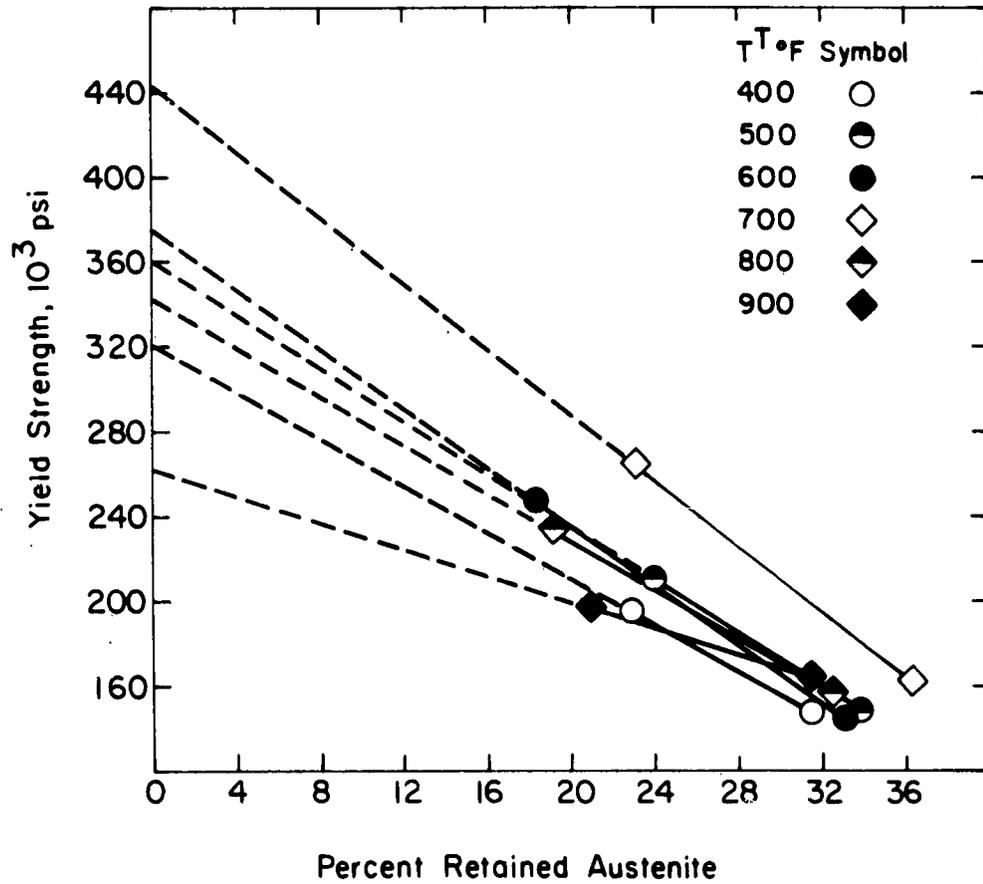


Fig.25- Yield strength of H11 as a function of percent retained austenite for various tempering temperatures in process AB. ( $T^A=1850^\circ\text{F}$ ,  $T^D=900^\circ\text{F}$  deformation = 50%  $T^B=550^\circ\text{F}$ ,  $t_f = 4$  hours,  $T^Q = \text{RT or LN}$ ).

370,000 psi, and an elongation of 3%. The percent reduction in area remains at approximately 25% for deformations up to 10%, and then gradually decreases to a level of 15% for the higher reductions.

The amounts of strengthening attributable to the deformation of the tempered bainite and to the retempering at 550°F were determined by testing specimens without retempering (i. e., with  $T^{T2} = RT$ ) and the results are shown in Fig. 26. An increment in yield strength of approximately 40,000 psi is produced by the retempering, and is independent of the extent of deformation in the range of 4 to 50%.

The effect of straining 10 percent and subsequent retempering on the microstructure of tempered bainite is illustrated by the light micrographs of Fig. 27. The structure in Fig. 27a consists essentially of tempered bainite and retained austenite. Straining 10 percent decreases the amount of retained austenite (Fig. 27b) and subsequent retempering appears to eliminate the austenite almost entirely (Fig. 27c). The straining also seems to cause some re-resolution of the carbides, and the subsequent retempering results in a precipitation of fine carbides. Both of these structural changes undoubtedly contribute to the 40,000 psi increment in yield strength, but the relative importance has not been ascertained as yet.

The effect of deformation on tempered martensite (process MS) will be discussed in section VC-1.

### 3. 2 Pretempering Temperature ( $T^{T1}$ ) and Retempering Temperature ( $T^{T2}$ )

The influence of pretempering temperature on the mechanical properties of H11 subjected to process BS is illustrated in

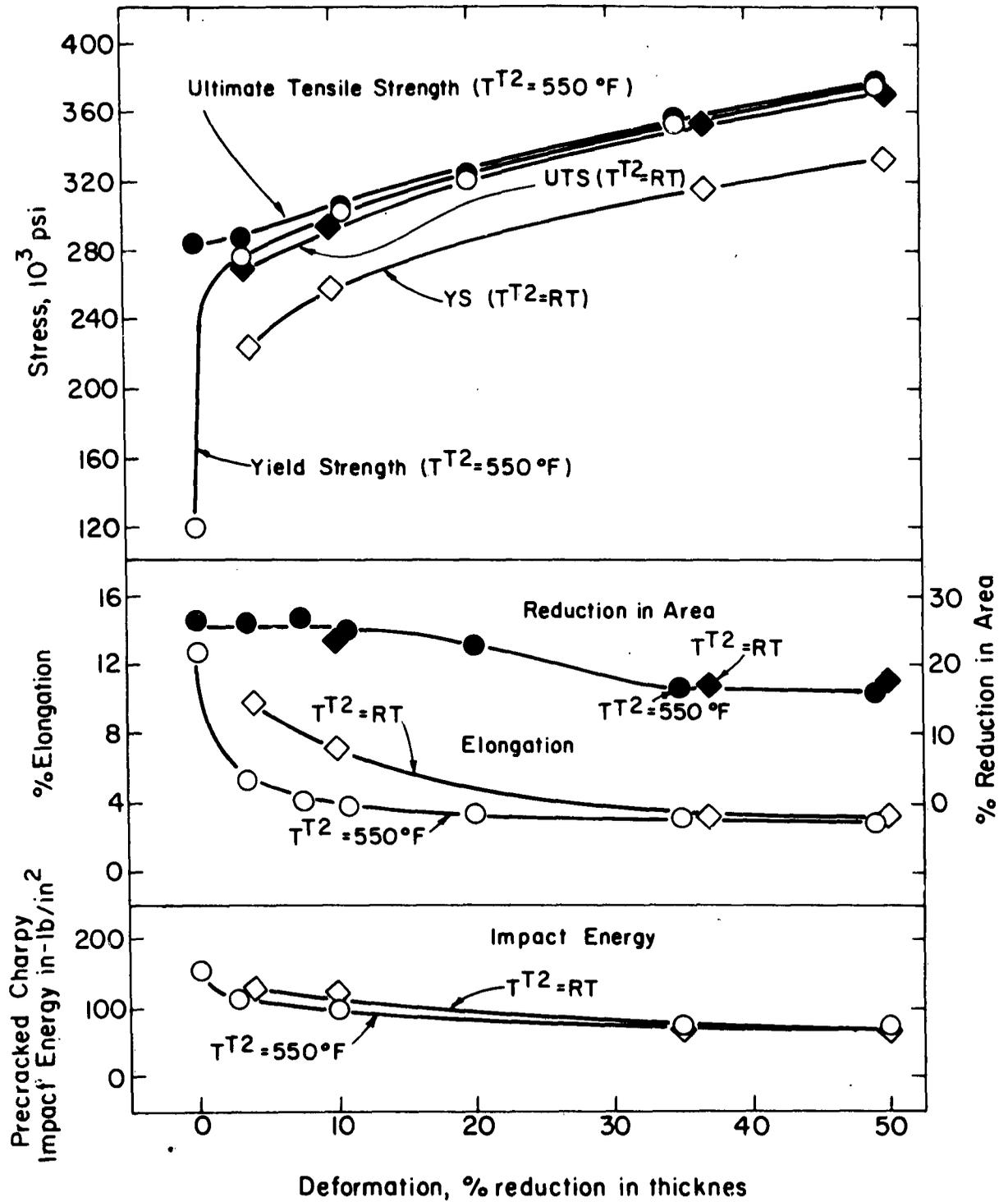


Fig.26-Mechanical properties of H11 as a function of percent reduction of thickness for process BS. ( $T^A = 1850^\circ\text{F}$ ,  $T^B = 550^\circ\text{F}$ ,  $t_f = 4$  hours,  $T^Q = \text{RT}$ ,  $T^{T1} = 550^\circ\text{F}$ ,  $T^{T2} = \text{RT}$  or  $550^\circ\text{F}$ ).



(a)



(b)



(c)

Fig. 27 - Light Micrographs of H11 Bainite Transformed to Bainite at 550°F (4 hrs.) and (a) Tempered at 550°F, (b) Pre-tempered at 550°F and Strained 10% and (c) Pretempered at 550°F, Strained 10% and Retempered at 550°F. Etchant Modified Picral. Magnification 2000X

Fig. 28 for deformations of 10 and 35% reduction in thickness. The trends in mechanical properties are similar for both amounts of deformation. The strength is unaffected by variations in  $T^{T1}$  from RT to 800°F. However, a marked strengthening peak occurs between 900 and 1100°F. A deformation of 35% with a  $T^{T1}$  of 550°F develops a yield strength of 350,000 psi, but this is increased to 395,000 psi by employing a  $T^{T1}$  of 1000°F. The strength maxima are accompanied by minima in elongation, reduction in area and impact energy. The precipitation of alloy carbides on tempering in the vicinity of 1000°F considerably enhances the strengthening obtained by strain-tempering this steel. As indicated by the virtual coinciding of the corresponding yield strength and ultimate tensile strength values in this series it appears that the uniform elongation values are negligible.

The experiments on varying the retempering temperature ( $T^{T2}$ ) are summarized in Fig. 29. A striking effect was found for 10% deformation and  $T^{T2}$  of 1000°F. The elongation increases from less than 4% to 14%, and the yield and ultimate strengths are 285,000 psi and 300,000 psi respectively.

### 3. 3 Effect of Retained Austenite

The influence of strain tempering on retained austenite and other structural features of H11 bainite formed at 550°F has been investigated to some extent. It is found that H11 tempered at 550°F and strained 10 percent by rolling causes a slight decrease in retained austenite from 34.4 to 31.1 percent with a large increase in yield strength from 119,000 psi to 257,000 psi. Upon aging this strained

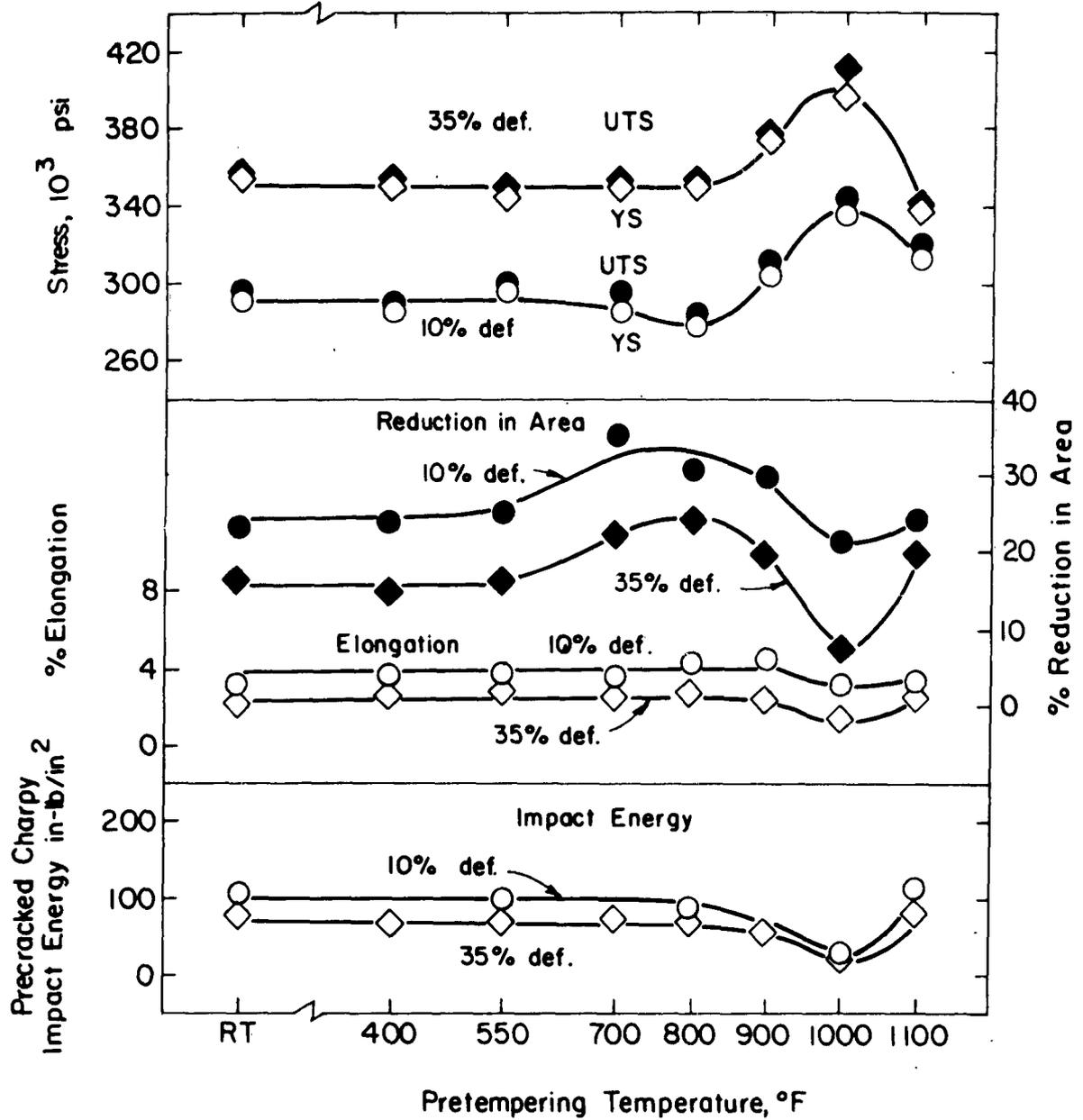


Fig. 28 - Mechanical properties of H11 as a function of pretempering temperature ( $T^{T1}$ ) for process BS. ( $T^A=1850^\circ\text{F}$ ,  $T^B=550^\circ\text{F}$ ,  $t=4$  hours,  $T^Q=RT$ , deformation = 10% or 35%  $T^{T2}=550^\circ\text{F}$ ).

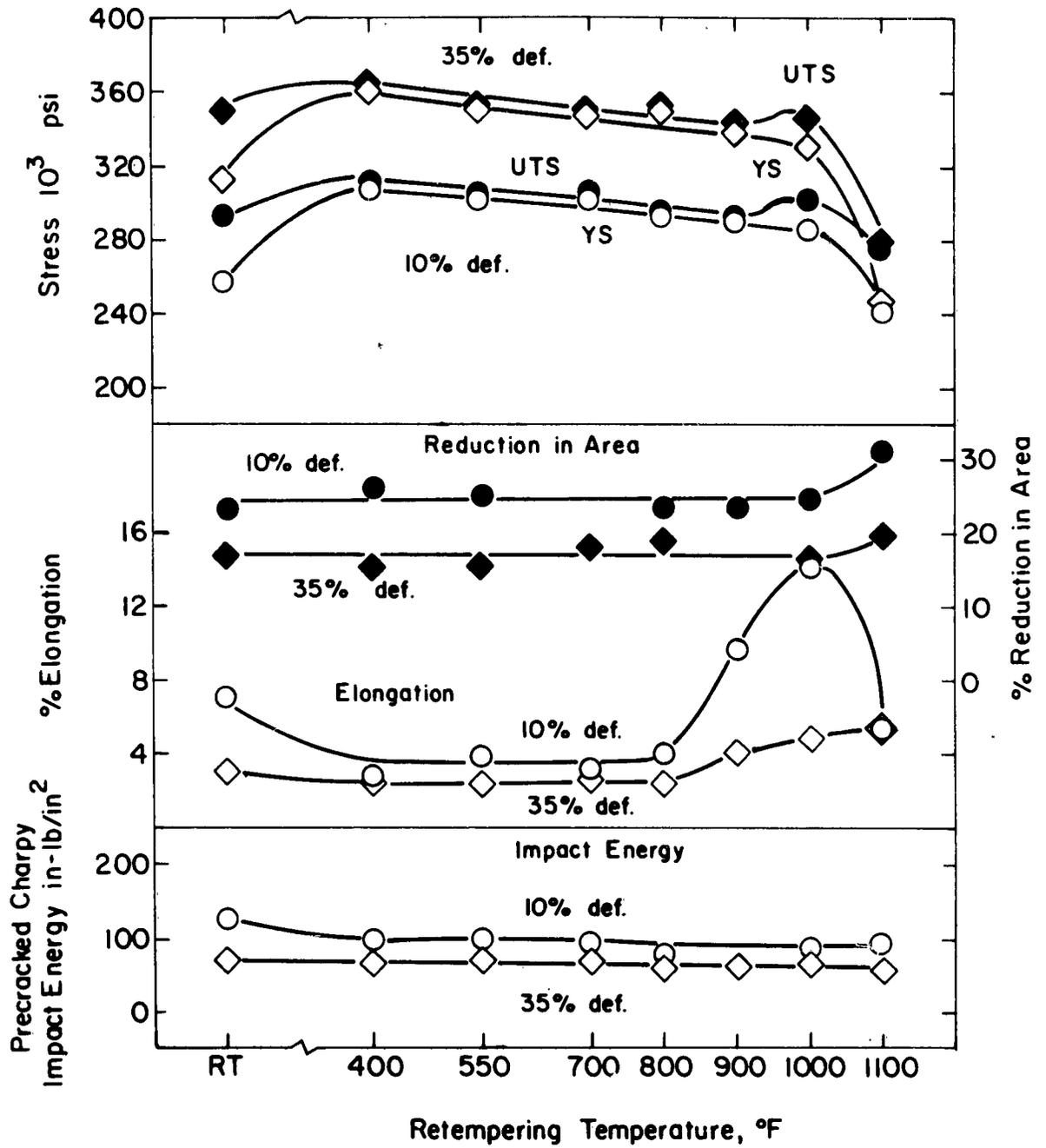


Fig.29- Mechanical properties of H11 as a function of retempering temperature ( $T^{T2}$ ) for process BS. ( $T^A=1850^\circ\text{F}$ ,  $T^B=550^\circ\text{F}$ ,  $t_f=4$  hours,  $T^Q=RT$ ,  $T^{T1}=550^\circ\text{F}$ , deformation = 10% or 35%).

structure, the austenite content decreases to 15.8 percent and the strength increases to 301,000 psi. From these initial data, it appears that the straining (with room temperature aging) causes most of the strengthening, but that the retempering at 550°F improves the yield strength even further by converting a large amount of austenite. It is evident that the austenite is made less stable by straining. Therefore, in order to confirm the relative importance of the straining and retempering steps in contributing to the final strength of bainite, the possible conversion of austenite during tensile testing must be explored.

#### 3.4 Combinations of Thermomechanical Treatments (Processes ABS and AMS)

It has been shown that one can obtain strength improvements in H11 by deformation of austenite before transformation to either bainite or martensite, or by deformation applied after transformation. The possibility of combining these deformation processes in order to achieve even greater strength levels was considered. As a result, experiments were performed to determine the total changes in properties as well as the change in each step of the combined treatments.

A comparison of the tensile properties of H11 bainite subjected to various thermal and thermomechanical treatments is given in Fig. 30. Treatments I through VI have been discussed individually in the preceding sections. Treatments VII and VIII (process ABS) demonstrate the strengthening advantage of combining the two deformation processes. A comparison of treatments VII and V shows that the strengthening obtained by bainite deformation is approximately additive to the strengthening increment obtained by austenite deformation prior to the bainitic reaction.

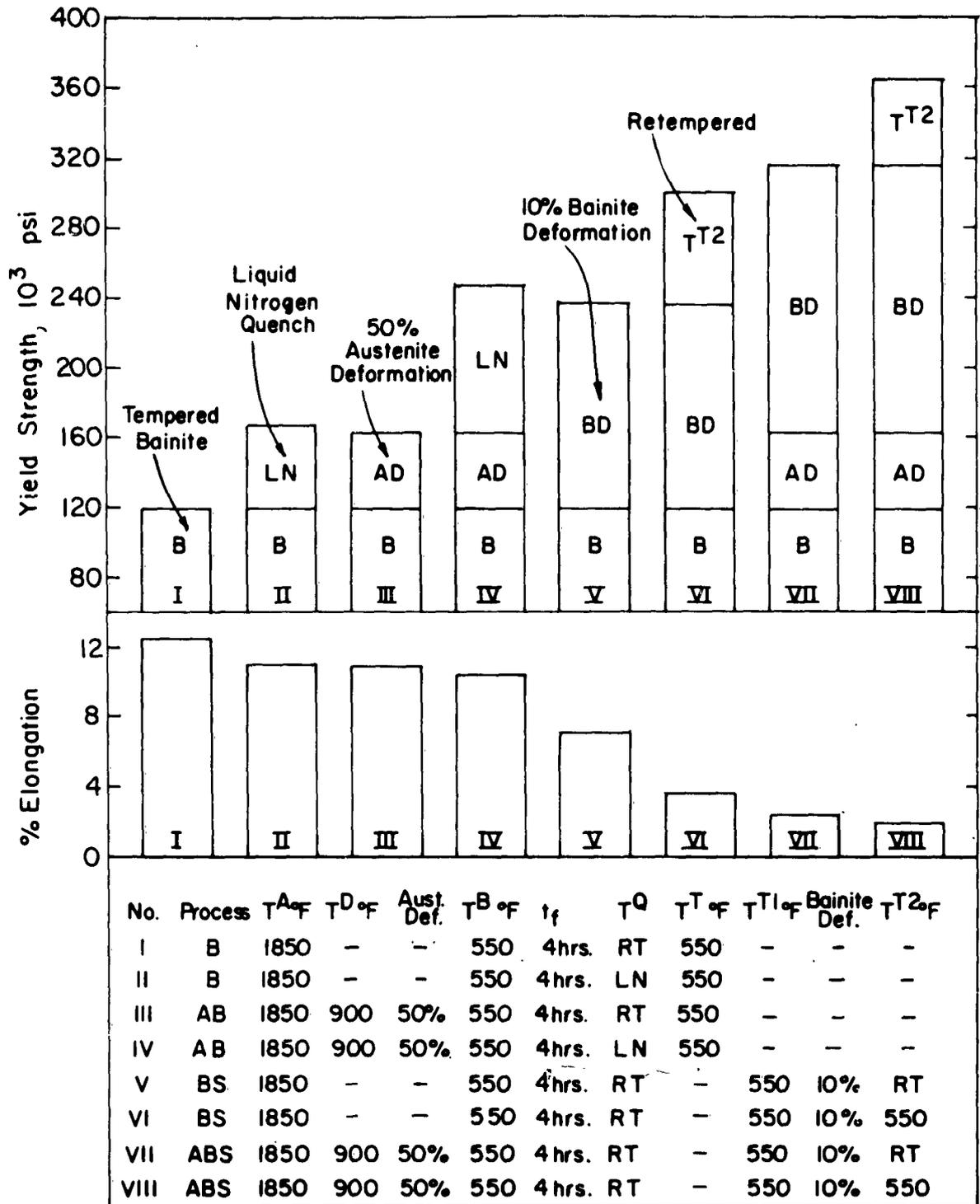


Fig.30-Comparison of tensile properties of H11 bainite subjected to various thermomechanical treatments.

As a matter of fact, the amount of strengthening obtained by 10% bainite deformation is not essentially altered by 50% prior deformation of the austenite. Similarly, retempering after the bainite-deformation step in process ABS produces a significant improvement in strength. It may be noted that LN treatments have no effect in processes involving deformation of the tempered bainite (treatments V through VIII). Process ABS, including retempering at 550°F, results in a yield strength of 365,000 psi which is the highest value attained in this series.

In general, it appears that the elongation decreases as the yield strength increases. However, process AB with a LN quench (treatment IV) shows an improved combination of properties. For this situation, the yield has been increased to 250,000 psi while the elongation has decreased only slightly to a value of 10%.

Fig. 31 depicts the strengthening gains obtained by combining austenite deformation with subsequent straining of martensite. As in the work on bainite, it is found that the strengthening effects from processes AM and MS are essentially additive. Process AMS (treatment IV) develops an exceptionally high yield strength of 390,000 psi; however, the elongation is only 1.5%.

C. Mechanical Properties of 4350 Steel Subjected to Thermomechanical Treatments

1. Bainite - No Deformation (Process B)

The effect of the bainite-reaction temperature ( $T^B$ ) on tensile properties of 4350 steel is shown in Fig. 32. The strength values decrease with increasing reaction temperature up to a temperature of

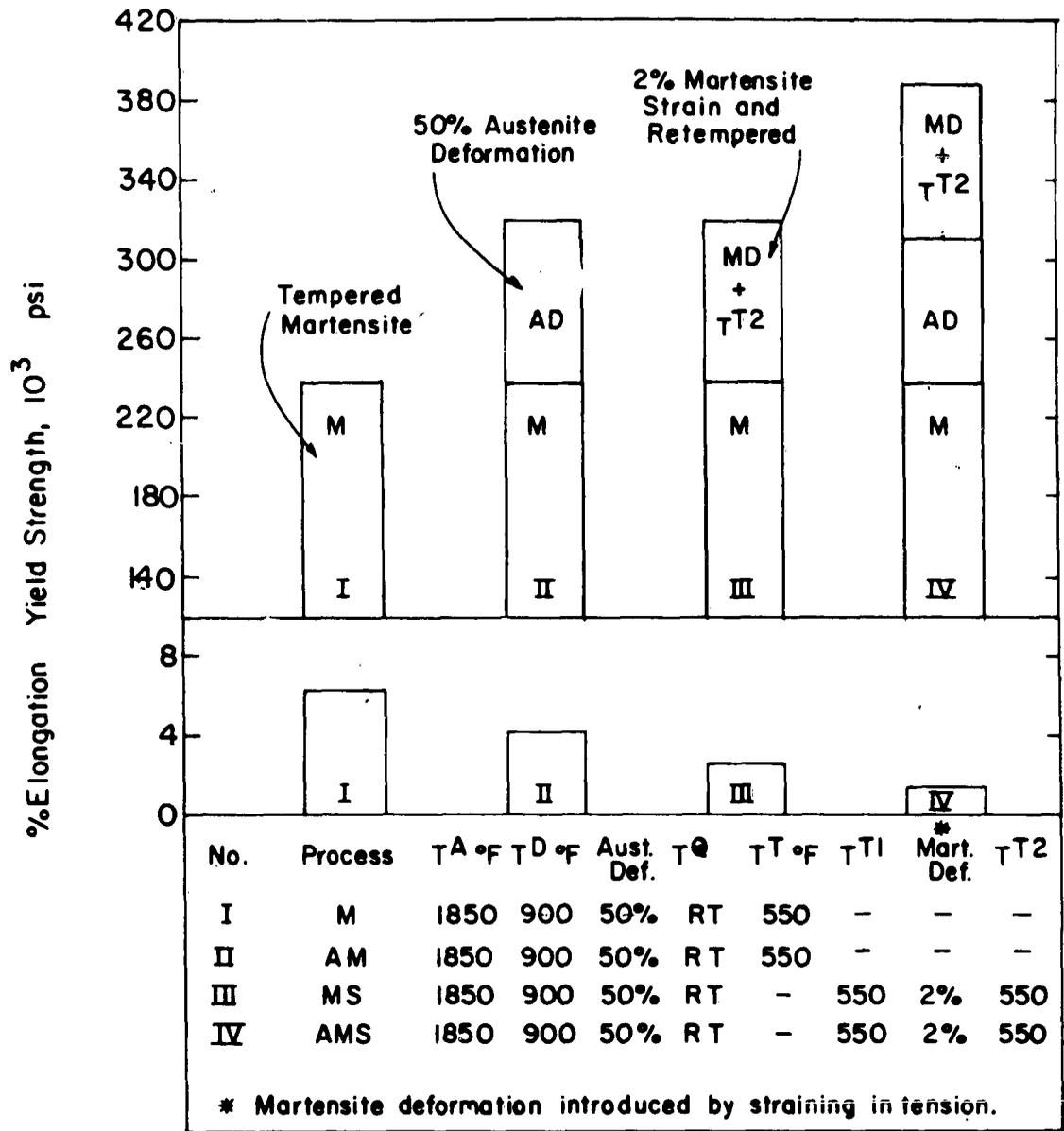


Fig.31 - Comparison of yield strength of H11 martensite subjected to various thermomechanical treatments.

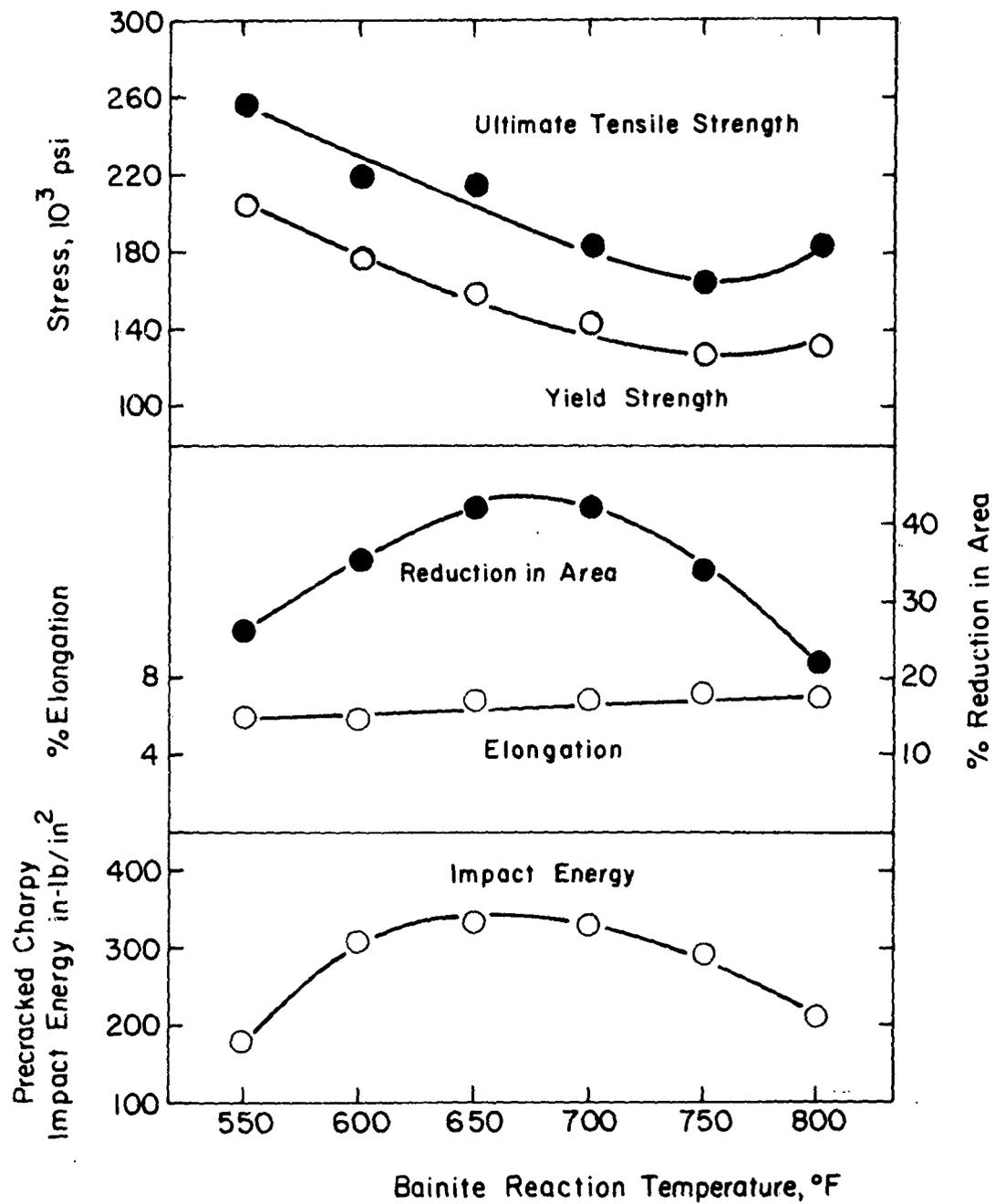


Fig. 32 - Mechanical properties of 4350 as a function of bainite reaction temperature for process B ( $T^A=1550^\circ\text{F}$ ,  $t_f=4$  hours  $T^T=550^\circ\text{F}$ ).

750°F, whereas the reduction in area and impact-energy values go through a maximum between 650 and 700°F. The % elongation is essentially constant over the entire range of  $T^B$  studied. At a  $T^B$  of 800°F, the reaction is still incomplete after 4 hours (Fig. 3) which allows some martensite to form on subsequent quenching from  $T^B$  with a resultant increase in strength as compared to a  $T^B$  of 700°F. Metallographic observations indicate a coarsening of the bainite occurs with increasing  $T^B$ .

The effect of tempering temperature ( $T^T$ ) on the bainite formed at 550°F is shown in Fig. 33. Small peaks in both yield and tensile strength occur at 600°F. For  $T^T$  values above 700°F, the yield strength drops below 200,000 psi and there is a large increase in impact energy.

## 2. Austenite Deformation Prior to Transformation (Processes AB and AM)

In contrast to the results obtained on H11, process AB applied to the 4350 steel results in only small improvements in strength, approximately 5,000 to 10,000 psi, over process B. This is the case even though  $T^B$  is varied from 550 to 750°F ( $T^T = 550°F$ ) and  $T^T$  is varied from RT to 1100°F ( $T^B = 550°F$ ). Austenite deformation prior to bainite formation produces a slight decrease in both elongation and impact energy.

The strengthening developed by process AM (ausforming) as a function of  $T^T$  is shown in Fig. 34. The tensile properties for process M are included in this figure in order to indicate the changes attributable to the austenite deformation. The yield strength increment decreases from 40,000 to 22,000 psi as  $T^T$  is increased from 400 to 1100°F.

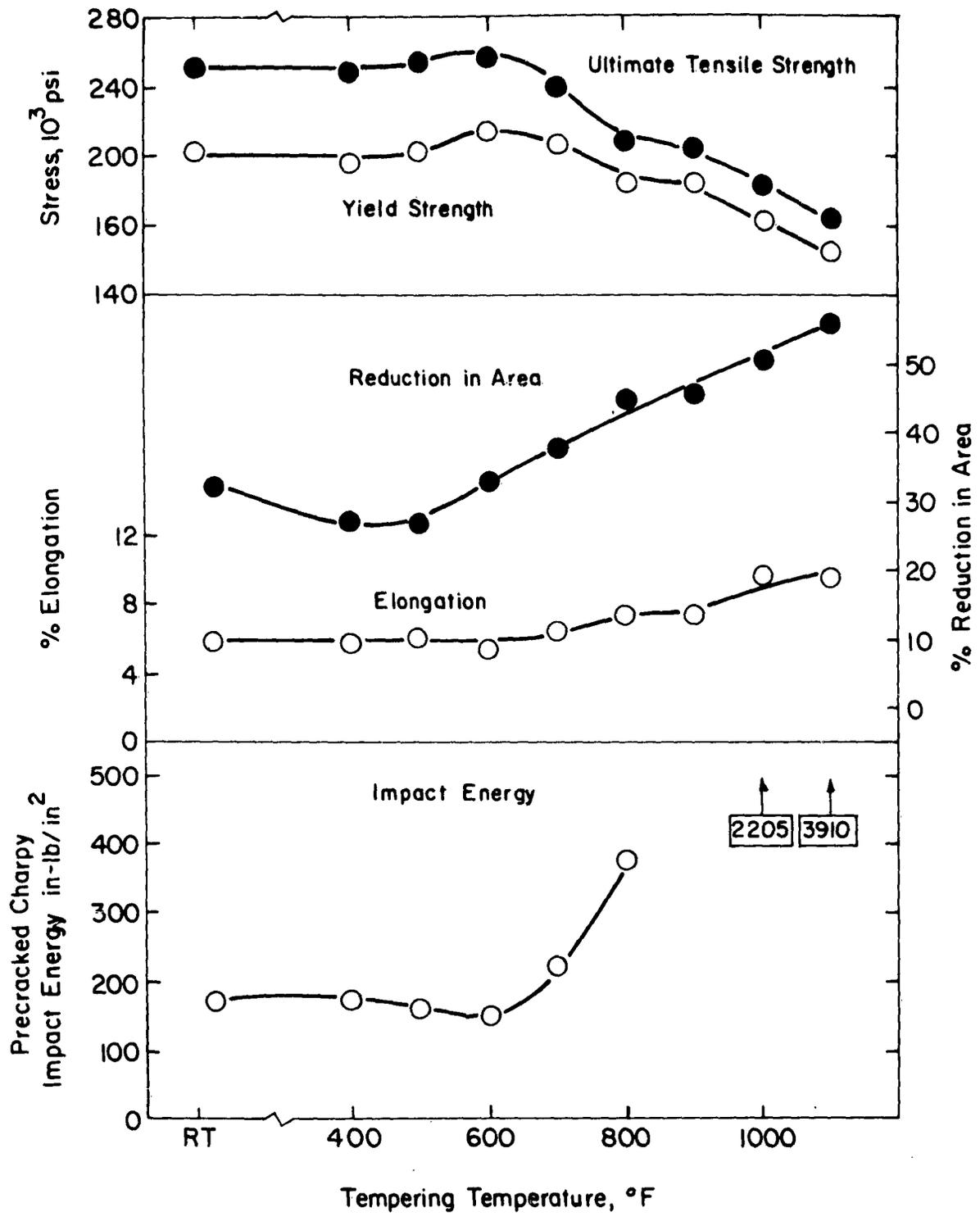


Fig. 33 - Mechanical properties of 4350 as a function of tempering temperature for process B. ( $T^A=1550^\circ\text{F}$ ,  $T^B=550^\circ\text{F}$ ,  $t_f = 4$  hours,  $T^Q = \text{RT}$ ).

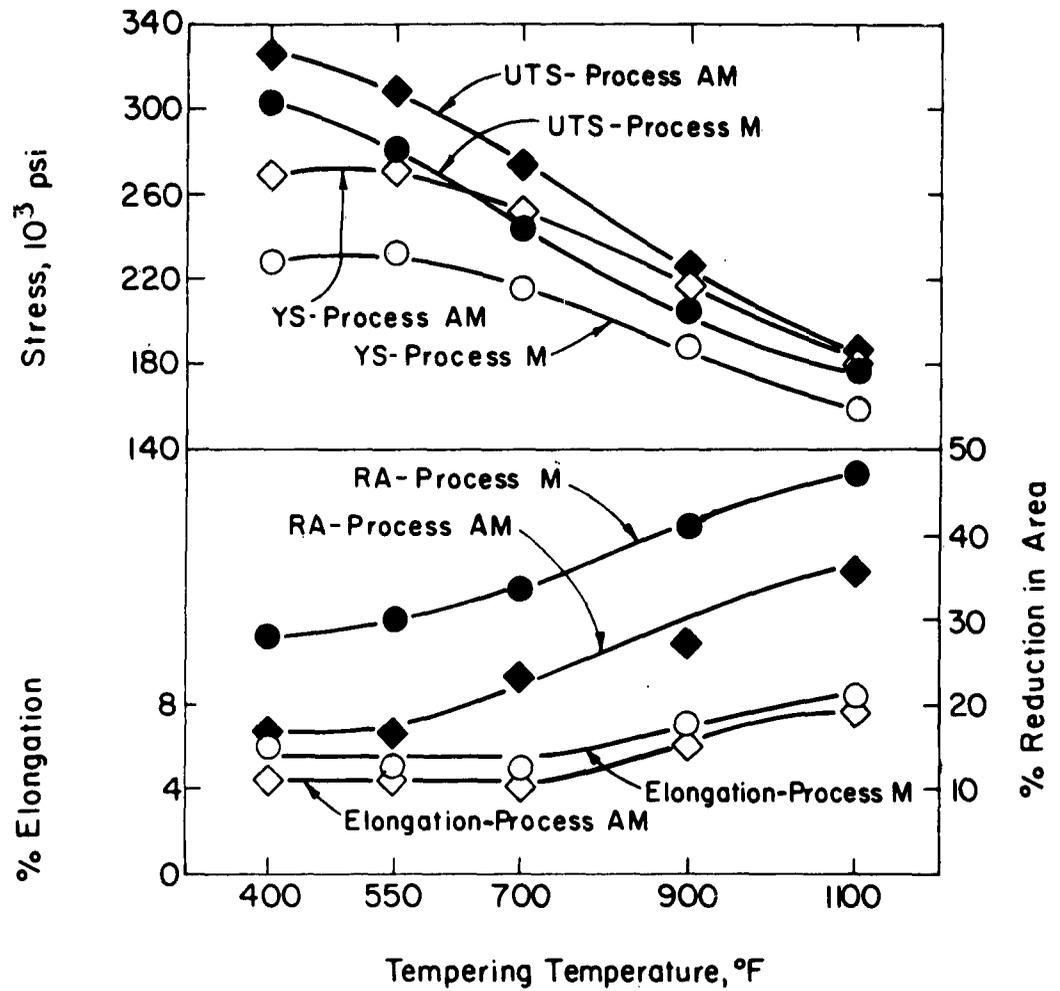


Fig. 34.-Tensile properties of 4350 as a function of tempering temperature ( $T^T$ ) for process M and AM. ( $T^A=1550^\circ\text{F}$ ,  $T^D=1050^\circ\text{F}$ , deformation = 5%,  $T^Q=RT$ ).

Some experiments were also performed to determine the influence of austenitizing temperature ( $T^A$ ) on processes M and AM. The results are summarized in Table 3. Changing  $T^A$  from 1550 to 2100°F has little effect on the levels of the mechanical properties obtained in process AM, which is in agreement with the findings of Justusson and Schmatz (7) on a 0.32% C, 3.0% Cr, 1.5% Ni steel. This is further confirmation that austenitic grain size, which determines the maximum martensitic plate size, does not control the strength of ausformed low-alloy steels. In contrast, the strength values as achieved by regular hardening and tempering (process M) decreases as the austenitizing temperature is raised.

Table 3

Effect of Austenitizing Temperature on the Tensile Properties of 4350 for Processes M and AM

Process	$T^A$ , °F	Yield Strength, psi	Ultimate Tensile Strength, psi	% elongation
M ( $T^Q = RT, T^T = 550^\circ F$ )	1550	232,000	281,000	4.9
	1700	229,000	278,000	3.7
	1900	218,000	263,000	4.2
	2100	210,000	254,000	4.7
AM ( $T^D = 105^\circ F, aust. def. = 5.0\%$ ) ( $T^Q = RT, T^T = 550^\circ F$ )	1550	270,000	308,000	4.5
	1700	265,000	308,000	3.3
	1900	263,000	307,000	3.1
	2100	266,000	306,000	3.9

3. Deformation of Bainite and Martensite (Processes BS and MS)

3.1 Bainite-Reaction Temperature ( $T^B$ )

The mechanical properties of strain-tempered 4350 as a function of  $T^B$  are shown in Fig. 35. As compared with process B, an increment in yield strength of approximately 50,000 psi occurs over the

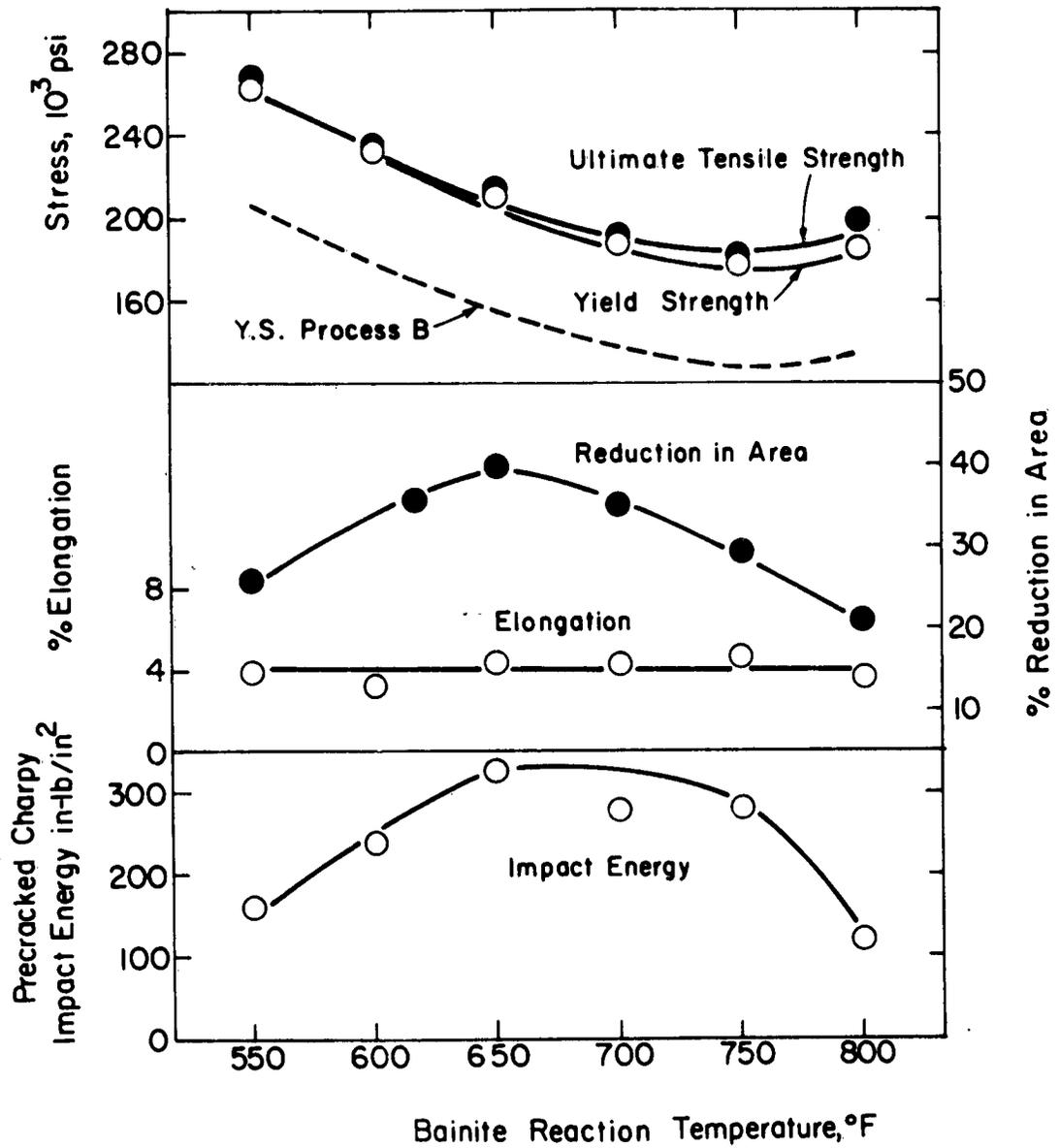


Fig. 35 - Mechanical properties of 4350 as a function of bainite reaction temperature ( $T^B$ ) for process BS. ( $T^A=1550^\circ\text{F}$ ,  $t_f=4$  hours,  $T^Q=\text{RT}$ ,  $T^{T1}=550^\circ\text{F}$ , deformation = 10%,  $T^{T2}=550^\circ\text{F}$  ).

entire range of  $T^B$  from 550 to 800°F. The reduction in area and impact energy go through a maximum at about 650°F similar to process B (Fig. 32). The elongation has a constant value of 4% over the transformation temperature range investigated. There is no obvious difference between the responses of "upper" and "lower" bainites to this strain-tempering treatment.

### 3.2 Pretempering Temperature ( $T^{T1}$ ) and Retempering Temperature ( $T^{T2}$ )

The tensile properties of 4350 as a function of  $T^{T1}$  and  $T^{T2}$  for process BS are summarized in Figs. 36 and 37, respectively. The yield strength is maximum for a value of  $T^{T1}$  and  $T^{T2} = 550^\circ\text{F}$ . The yield-to-ultimate strength ratio is approximately unity when  $T^{T2}$  is 550°F. This ratio decreases slowly as  $T^{T2}$  is either decreased or increased from 550°F.

A higher yield strength level is obtained on retempering at high temperatures ( $T^{T2}$  up to 1100°F) as compared to tempering unstrained bainite (Fig. 33). For example, straining 10% and retempering at 900°F results in a yield strength as high as about 200,000 psi, (along with a reduction in area of 48%, and an elongation of 7%).

### 3.3 Percent Deformation

The effect of the degree of deformation on processes BS and MS for 4350 are illustrated in Fig. 38. The strain-tempering curves of H11 bainite and martensite are included for comparison purposes. In all cases, conspicuous strengthening is found as a result of small deformations of the order of 3 percent. For deformations of 5 to 50

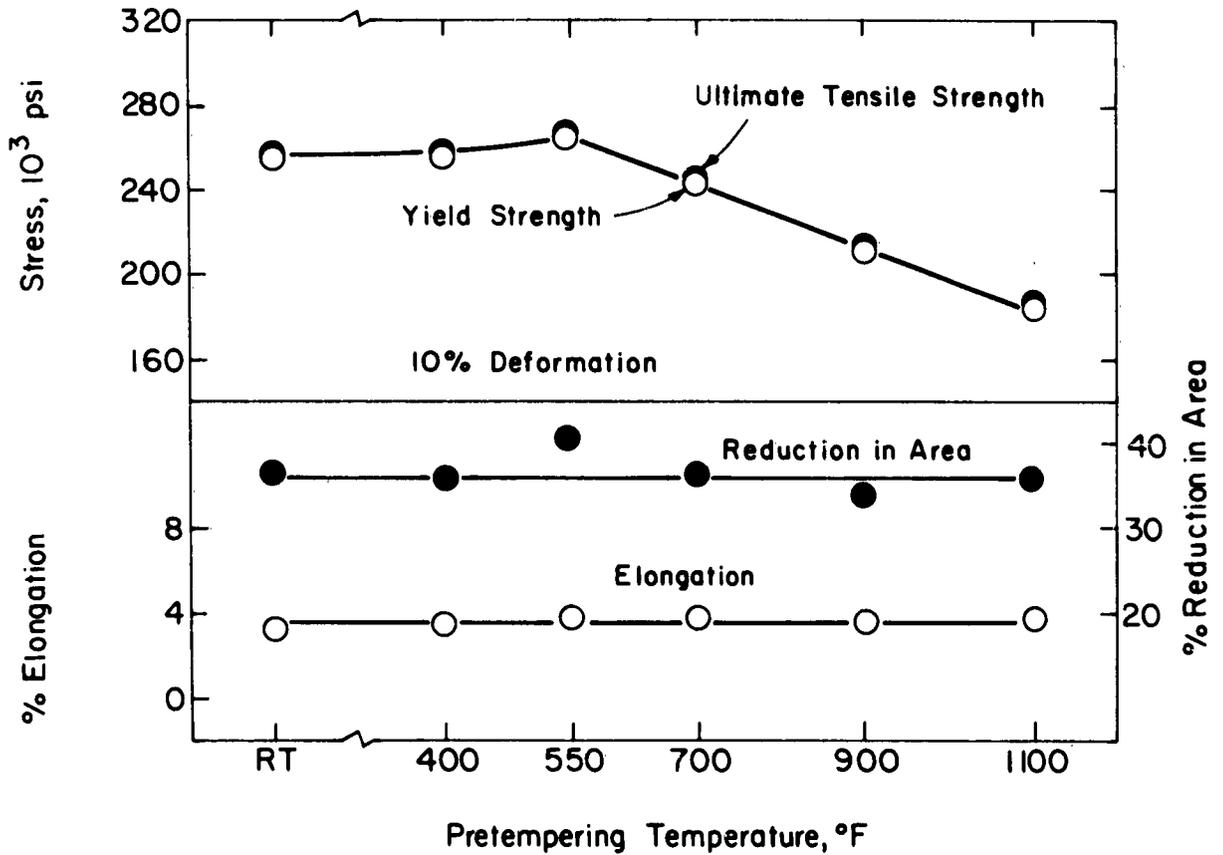


Fig.36 - Tensile properties of 4350 as a function of pretempering temperature ( $T^{T1}$ ) for process BS. ( $T^A=1850^\circ\text{F}$ ,  $T^B=550^\circ\text{F}$ ,  $t_f=4$  hours,  $T^Q=RT$ , deformation = 10%,  $T^{T2}=550^\circ\text{F}$ ).

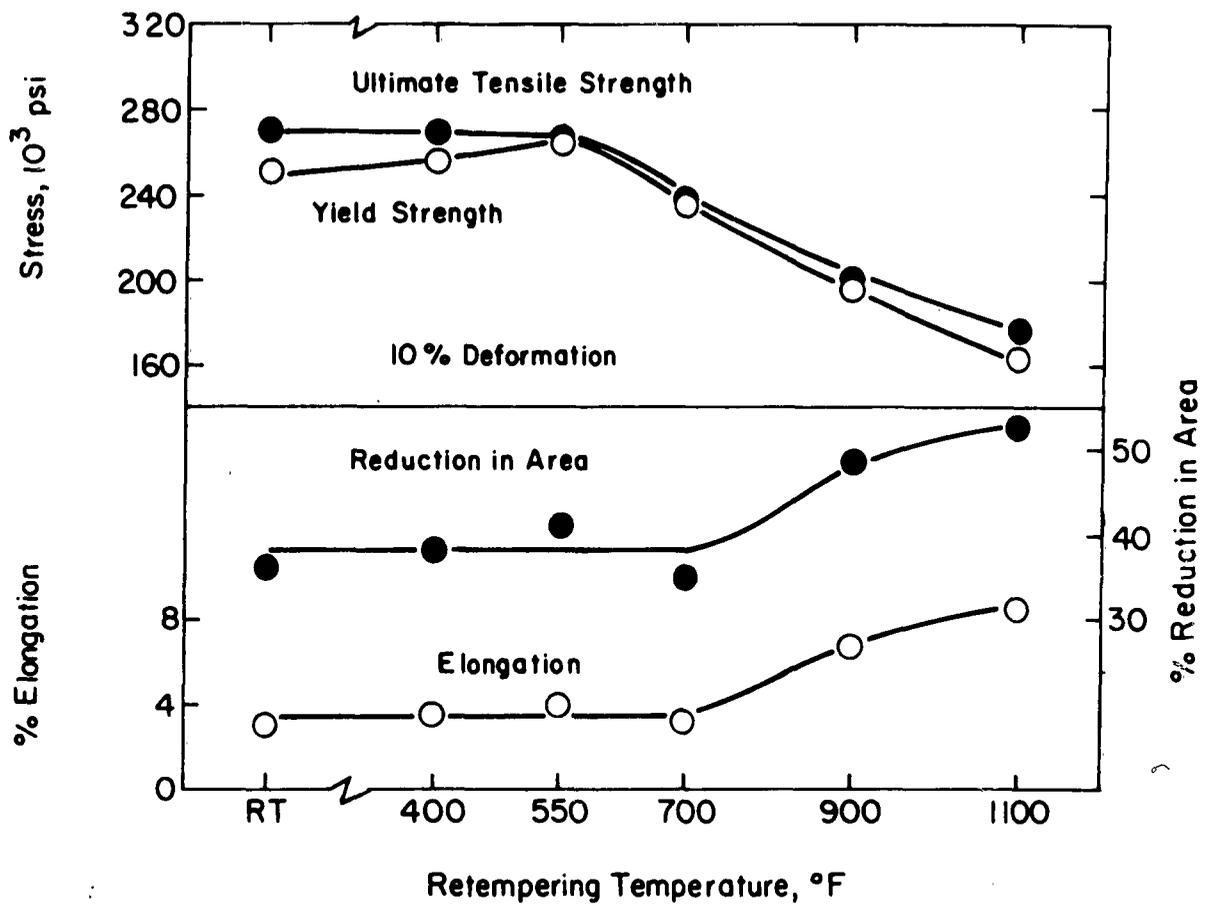


Fig.37 - Tensile properties of 4350 as a function of retempering temperature ( $T^{T2}$ ) for process BS. ( $T^A=1550^\circ\text{F}$ ,  $T^B=550^\circ\text{F}$ ,  $t_f=4$  hours,  $T^Q=\text{RT}$ ,  $T^{T1}=550^\circ\text{F}$ , deformation = 10%).

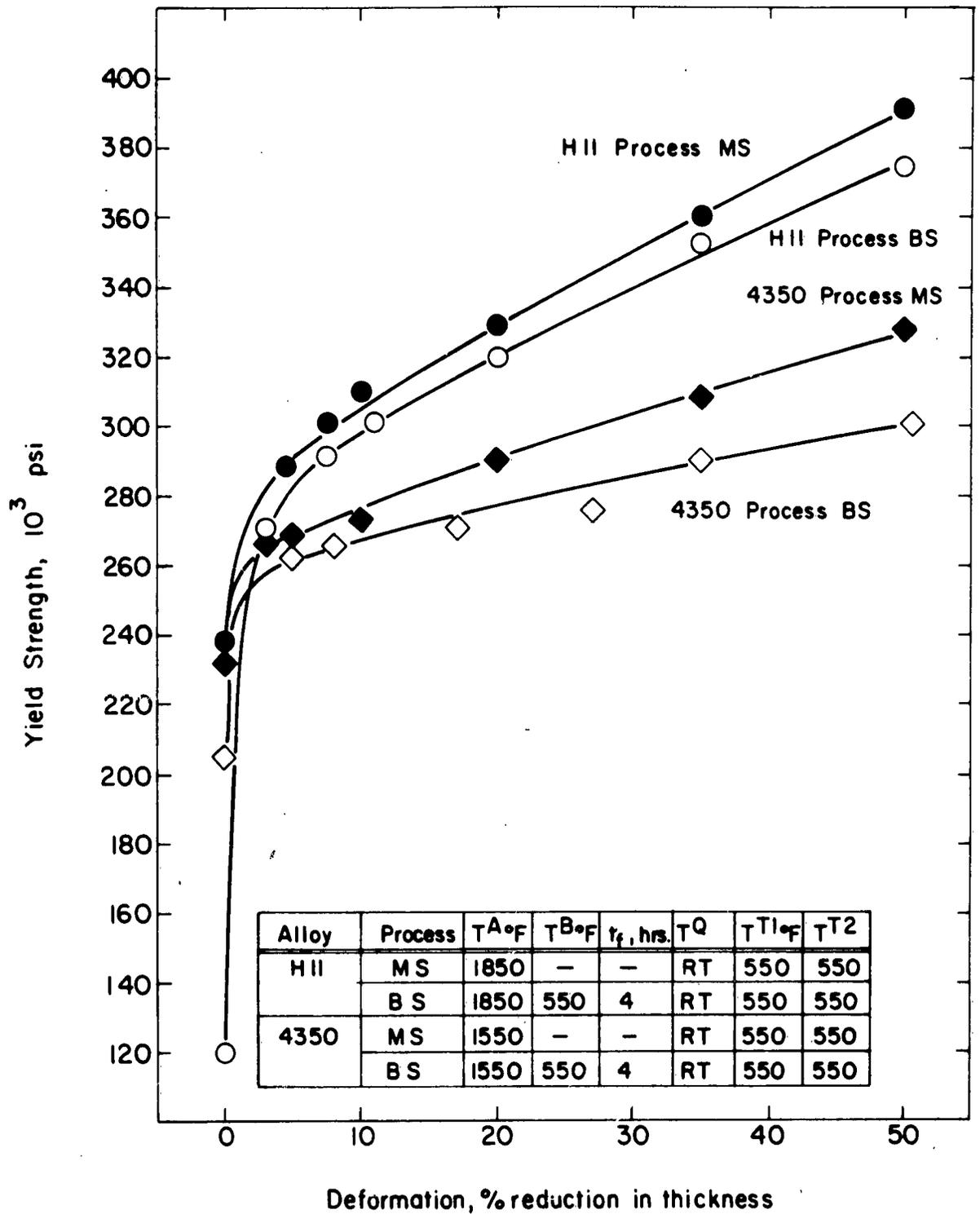


Fig.38 -Yield strength of 4350 and H11 as a function of percent deformation for processes BS and MS.

percent, both materials display an essentially linear increase in strength with increasing deformation.

Bainite has a greater strengthening response than does martensite at very low deformations, but the martensite strengthens more rapidly at higher amounts of deformation. The initial sensitivity of bainite to strain tempering is undoubtedly due to the retained austenite present, rather than to the bainite itself. The H11 structures strain temper to higher strength levels than do the 4350 structures. For any given amount of deformation, the process MS develops higher strength in each material than does process BS.

#### 4. Combinations of Thermomechanical Treatments (Processes ABS and AMS)

In view of the small strengthening improvement obtained by austenite deformation prior to bainite formation in the 4350 steel (process AB vs. process B), there seems to be no advantage to employing a combination treatment. In other words, one should obtain practically the same results for process ABS as for process BS in this material.

However, combination treatments can be applied advantageously to martensites in the 4350 steel. Fig. 39 shows the almost additive nature of the yield strength increments associated with each thermomechanical treatment. For example, treatments II and IV (processes AM and MS) produce yield-strength increments of 38,000 and 50,000 psi, respectively, and combining these treatments in treatment VI (process AMS) achieves an increment of 78,000 psi. Once again, it is observed that as the strength of martensite is improved the ductility tends to decrease.

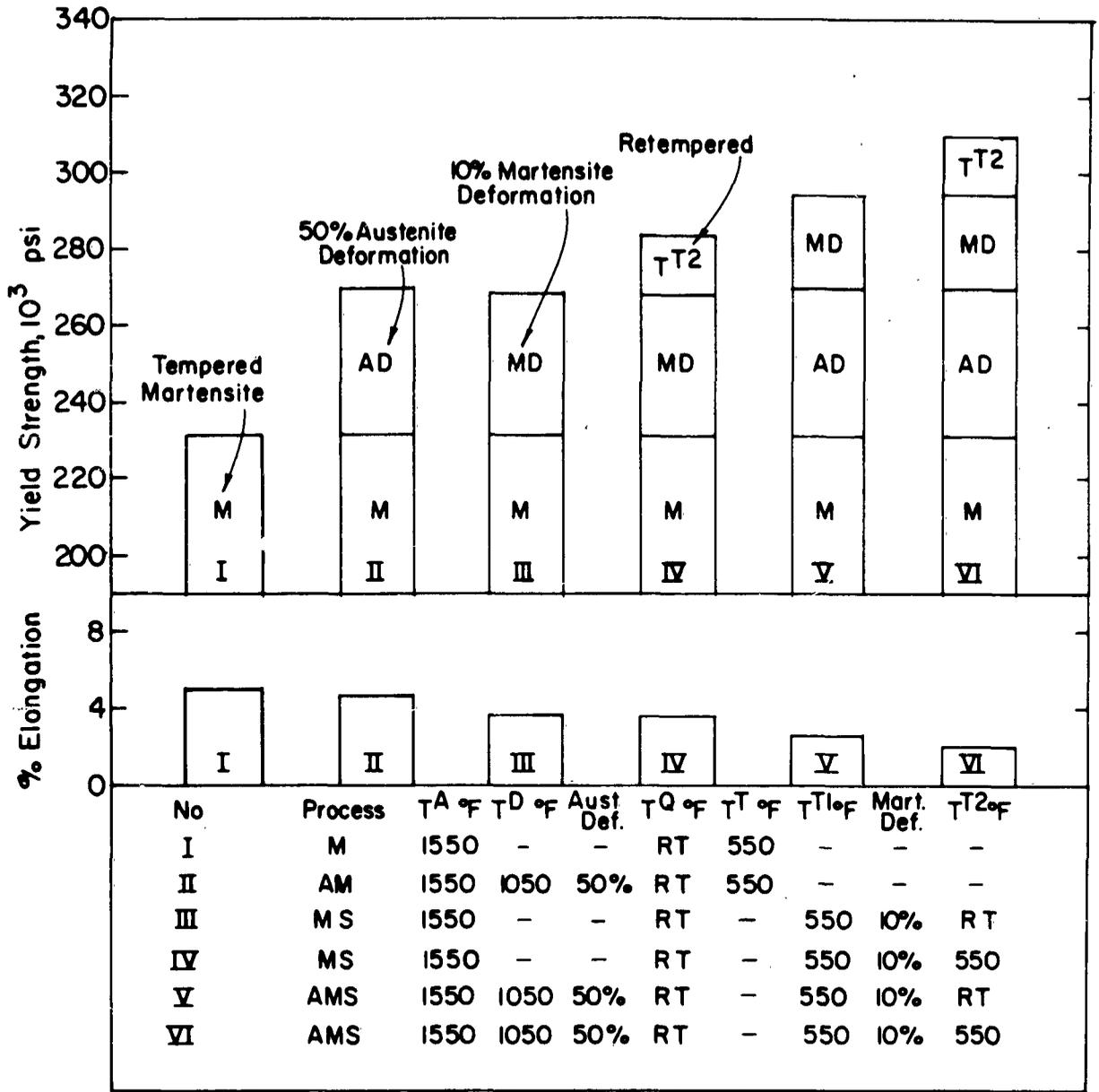


Fig.39- Comparison of yield strength at 4350 martensite subjected to various thermomechanical treatments.

## 5. Effect of Carbon Content on Process MS

Supplementary experiments were performed to explore the influence of carbon content on the strain-tempering response of 43XX martensites. Some data were available for carbon contents of 0.12 and 0.20 percent <sup>(8)</sup>, and carbon contents of 0.35 and 0.43 percent were added to the program. The results of these experiments involving process MS are presented in Fig. 40. It may be seen that the strength improvements for both straining and retempering increase with the carbon content.

### D. GENERAL COMPARISONS AND CONCLUSIONS

#### 1. General Comparisons

The response of H11 and 4350 to various thermal and thermomechanical treatments is shown in Figs. 41 and 42, respectively. These data provide a direct comparison between bainite and martensite after analogous processing conditions. The processes considered in these figures involve either: (1) no deformations, (2) austenite deformation, or (3) transformation-product deformation (strain tempering). Under these conditions, with a 550<sup>o</sup>F temper, it is seen that the martensites always have greater strengths than the bainites in both H11 and 4340. However, the bainites generally have superior ductility. In H11 the differences between martensitic and bainitic properties are most striking for treatments involving either no deformation or austenite deformation. In 4350, the ductility differences are small, but the strength variation is significant for all processes.

An interesting comparison may be made between austenite deformation processes and strain-tempering processes. Deformation of

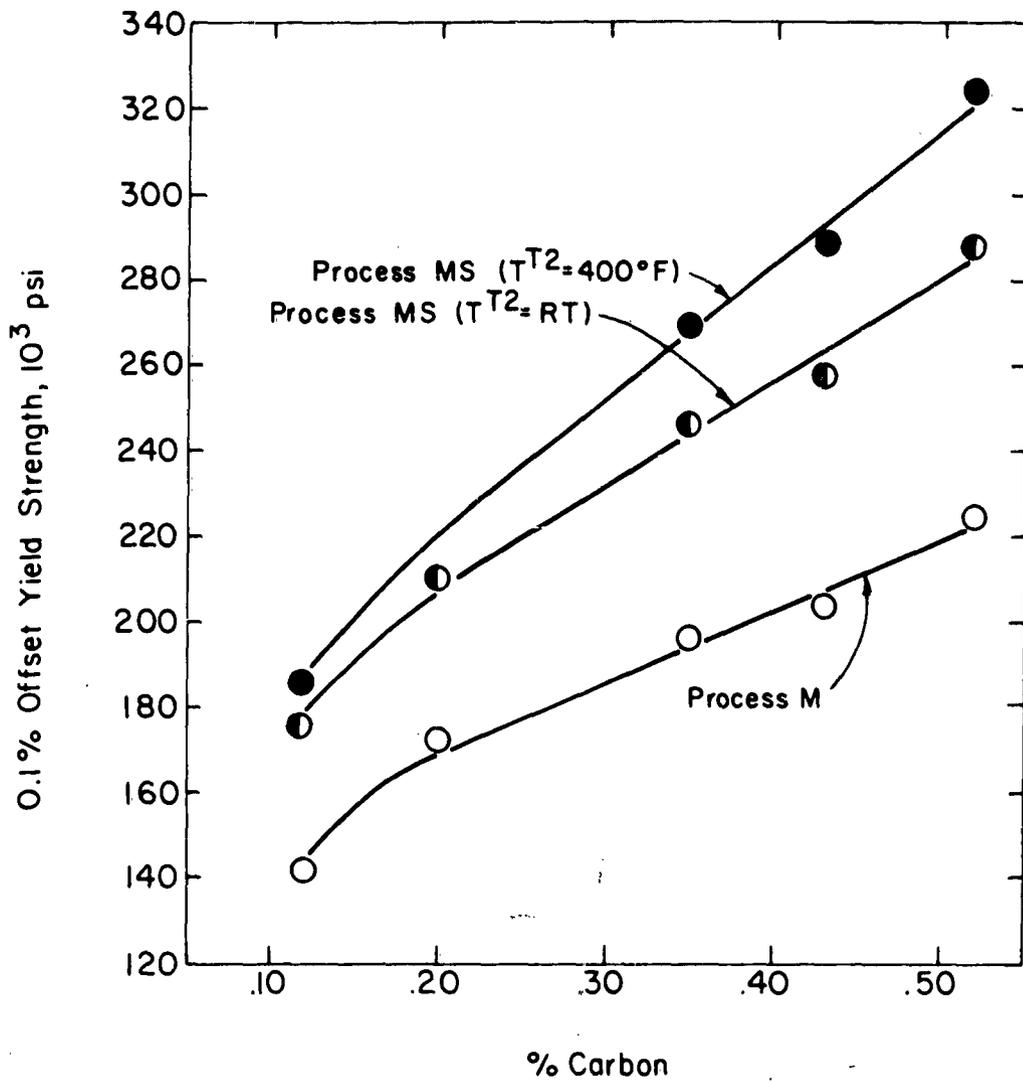


Fig. 40 - Influence of carbon content on strain tempering response of 43XX martensites. ( $T^T_1 = 400^\circ\text{F}$ , deformation = 10%). Data for 0.12 and 0.20 carbon from reference (8).

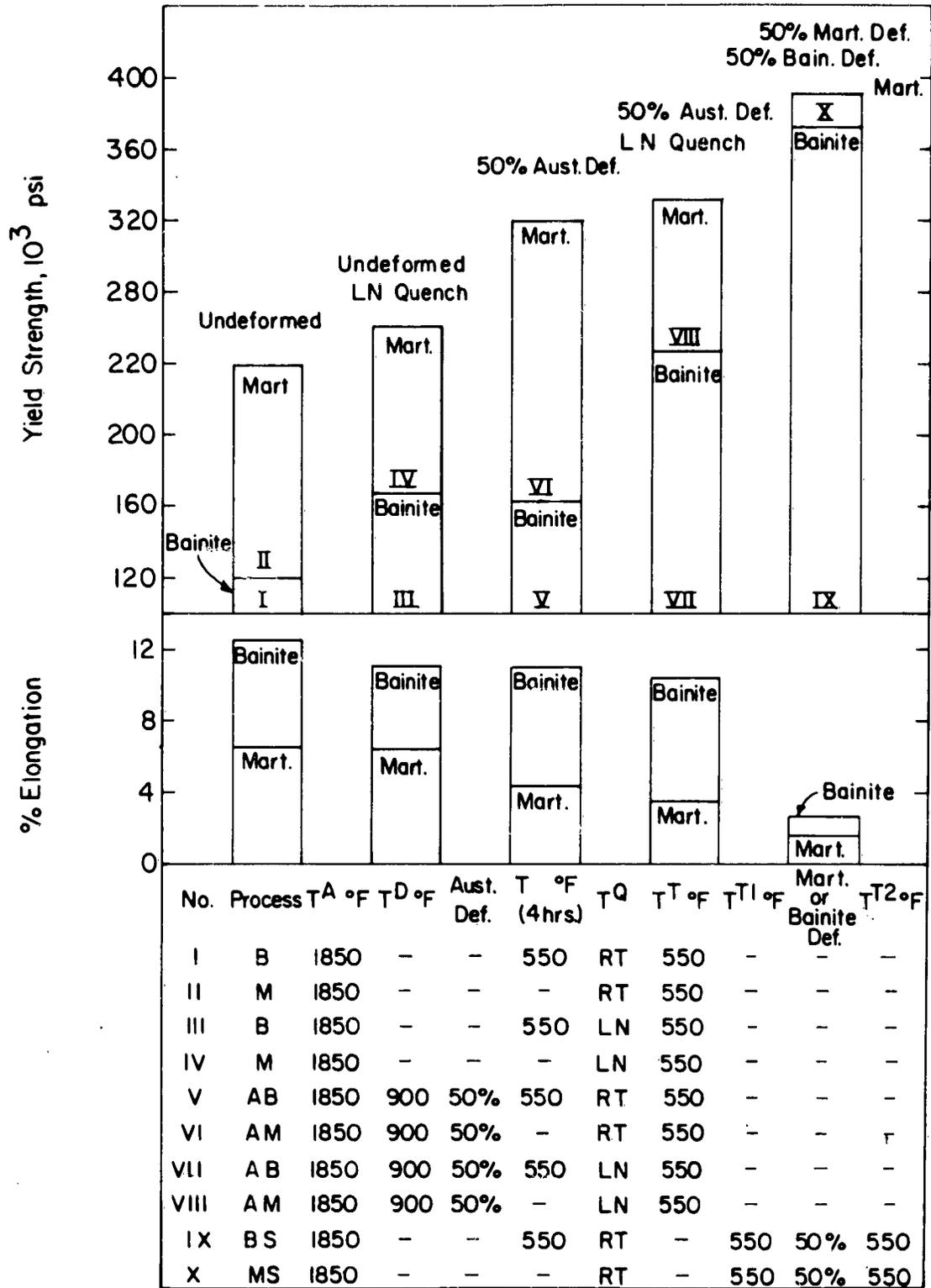


Fig. 41 - Comparison of the response of H11 bainite and martensite to various thermal and thermomechanical treatments.

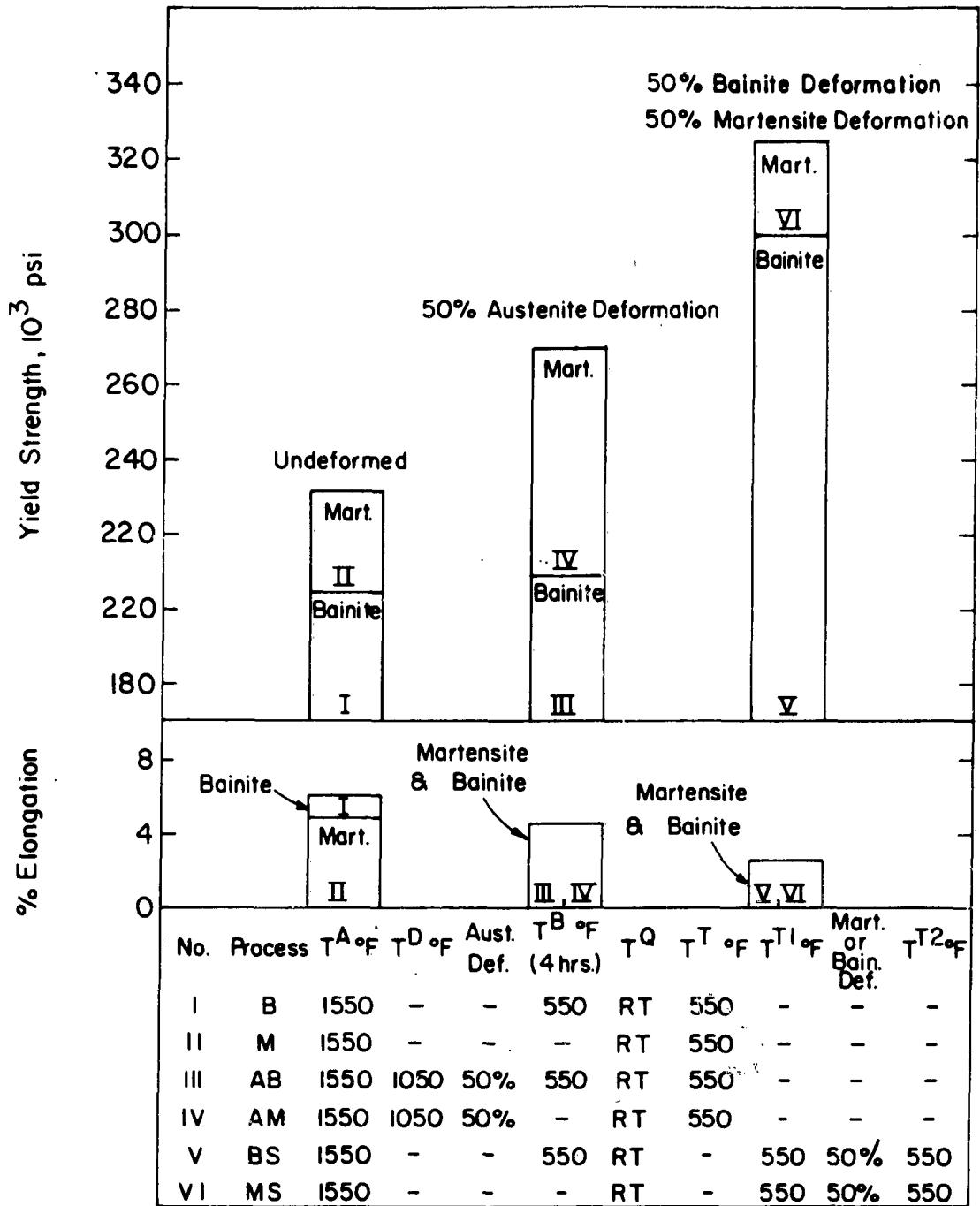


Fig.42 - Comparison of the response of 4350 bainite and martensite to various thermal and thermomechanical treatments.

either bainite or martensite (strain tempering) always develops much higher strengths for a given amount of deformation than does austenite deformation followed by transformation to either bainite or martensite. This may be seen for H11 in Fig. 41 by comparing treatments V and VI with IX and X, and for 4350 in Fig. 42 by comparison of treatments III and IV with V and VI. The strain-tempering treatments always produce lower ductilities than is obtained with the austenite-deformation processes.

Finally, a comparison may be made between the two steels investigated. H11 is found to be much more responsive than 4350 to all of the thermomechanical treatments investigated.

## 2. Conclusions

The results of this work can be summarized as follows:

1. The mechanical properties of H11 in the martensitic or bainitic condition are affected by refrigeration in liquid nitrogen when tempered at temperatures below 1000°F. This indicates the influence of retained austenite.
2. H11 transformed to bainite at 550°F exhibits a large secondary strengthening when tempered between 1000 and 1100°F. This effect is mainly attributable to the elimination of austenite which is retained in large amounts when bainite is formed.
3. After tempering at the highest tempering temperatures (1000 to 1200°F), H11 bainite has higher strength than does martensite, accompanied by equivalent ductility and toughness. Bainite tempered at 1100°F has a yield strength of approximately 235,000 psi as compared to approximately 212,000 for martensite tempered at the same temperature.

4. When only a very small amount of bainite is formed in H11 (transformation times up to 15 minutes), a marked decrease in fracture toughness occurs.

5. Austenite deformation in H11 of 50 percent depresses  $M_s$  by only  $5^\circ\text{F}$ , but significantly alters the shape of the martensite-range curve.

6. Austenite deformation in H11 results in considerable strengthening of the subsequently formed bainite, and refrigeration in liquid nitrogen is then found to have a large effect on tempering up to  $900^\circ\text{F}$ , but not at  $1000^\circ\text{F}$  and above. A strengthening peak occurs on tempering at  $1000^\circ\text{F}$ , with bainitic strengths being higher than is obtained on similar tempering of ausformed H11.

7. H11 bainites and martensites exhibit a large response to strain-tempering treatments. Small deformations (up to 3 percent) produce pronounced strengthening effects. Retempering produces an essentially constant strength increment (40,000 psi) for bainites strained between 5 and 50 percent.

8. Maximum strengthening due to strain tempering of H11 bainites is obtained by pretempering at  $1000^\circ\text{F}$  and retempering at  $400^\circ\text{F}$ .

9. The increments in strength developed by deformation processes both before and after transformation to either bainite or martensite are found to be approximately additive when the individual processes are combined.

10. When bainite is formed in H11 from either undeformed or deformed austenite, the resulting structure contains approximately 35 percent retained austenite.

11. Deformation of austenite in H11 prior to the bainitic transformation makes the austenite less stable relative to subsequent refrigeration treatments. Similarly, deformation of retained austenite makes the austenite less stable relative to further tempering.

12. The influence of retained austenite on the yield strength of H11 is significant for tempering temperatures below 1000°F, but the effect is not proportional to the volume percent of austenite.

13. Austenite deformation in 4350 has little effect on the strength of subsequently formed bainite. However, such deformation does enhance the strength properties of martensite.

14. Both bainitic and martensitic structures in 4350 exhibit a large response to strain-tempering treatments. As in the case of H11, small deformations (up to 3 percent) result in marked strengthening effects.

15. Increasing the carbon content in 43XX steels produces increasing increments in strength due to strain tempering.

16. The application of deformation processes both before and after martensite formation in 4350 results in a strengthening which is approximately additive when the individual processes are combined.

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