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Effective use of Weld Metal Yield Strength for HY-Steels

National Materials Advisory Board

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This report reviews the presently available welding technology for high-yield-strength steel (HY-steel) systems from the viewpoint of suggesting where fabrication cost savings might be achieved without sacrificing weldment integrity or performance. The review involves: economic analysis of fabrication costs; present criteria for high strength steel weldment performance, both military and industrial;

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present understanding of the effects of hydrogen in high strength steel weldments; factors affecting weld metal mechanical properties; and factors affecting weldment mechanical performance.

The study concludes that maximum fabrication-cost savings can be achieved by the elimination of the presently required preheat and reduction of the post weld inspection schedules, particularly for noncritical joints. The committee focused on the technical feasibility of eliminating preheat that is commonly used to avoid hydrogen-induced weld cracking. A literature review indicates that both a reduction of the hydrogen content of the weld and residual weld stresses could reduce the tendency for hydrogen-induced cracking. Welding processes with very low hydrogen potential are available (GMAW and SAW) and recent work in the covered electrode area (SMAW) have been shown to produce similar very low weld-metal hydrogen contents. Residual stresses in welds are related to weld metal yield strength; therefore, the use of matching or undermatching weld metal would be advantageous. Codes and fabrication documents show that, with the exception of a few special cases, weld metal strengths are required to meet or surpass the base metal requirements, apparently based on tradition, with little experimental data for backup. Studies have shown that in some cases, depending on joint design, weldments with weld-metal strengths only 90 percent of those of the base metal have performed satisfactorily. Since toughness is more readily attained at lower strength levels (for a particular steel weldment system), any enhancement of toughness would tend to offset some reduction in weld metal strength as far as weldment performance is concerned. In addition, should lower-strength weld metals prove adequate in performance, weld metal deposition rates may be used that would provide possible cost saving.

The overall recommendation is that there should be a reconsideration of the weld-metal yield-strength requirement. Should undermatching (or even matching) strength be proven satisfactory, this lower strength level, in conjunction with improved low-hydrogen welding consumables and techniques, could lead to reduction or even elimination of hydrogen cracking; hence the need for preheating.

EFFECTIVE USE OF
WELD METAL YIELD STRENGTH
FOR HY-STEELS

Report of
The Committee on Effective Utilization of
Weld Metal Yield Strength

NATIONAL MATERIALS ADVISORY BOARD
Commission on Engineering and Technical Systems
National Research Council

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The report has been reviewed by a group other than the authors according to procedures approved by a Report Review Committee consisting of members of the National Academy of Sciences, the National Academy of Engineering, and the Institute of Medicine.

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ABSTRACT

This report reviews the presently available welding technology for high-yield-strength steel (HY-steel) systems from the viewpoint of suggesting where fabrication cost savings might be achieved without sacrificing weldment integrity or performance. The review involves: economic analysis of fabrication costs; present criteria for high strength steel weldment performance, both military and industrial; present understanding of the effects of hydrogen in high strength steel weldments; factors affecting weld metal mechanical properties; and factors affecting weldment mechanical performance.

The study concludes that maximum fabrication-cost savings can be achieved by the elimination of the presently required preheat and reduction of the post weld inspection schedules, particularly for noncritical joints. The committee focused on the technical feasibility of eliminating preheat that is commonly used to avoid hydrogen-induced weld cracking. A literature review indicates that both a reduction of the hydrogen content of the weld and residual weld stresses could reduce the tendency for hydrogen-induced cracking. Welding processes with very low hydrogen potential are available (GMAW and SAW) and recent work in the covered electrode area (SMAW) have been shown to produce similar very low weld-metal hydrogen contents. Residual stresses in welds are related to weld metal yield strength; therefore, the use of matching or undermatching weld metal would be advantageous. Codes and fabrication documents show that, with the exception of a few special cases, weld metal strengths are required to meet or surpass the base metal requirements, apparently based on tradition, with little experimental data for backup. Studies have shown that in some cases, depending on joint design, weldments with weld-metal strengths only 90 percent of those of the base metal have performed satisfactorily. Since toughness is more readily attained at lower strength levels (for a particular steel weldment system), any enhancement of toughness would tend to offset some reduction in weld metal strength as far as weldment performance is concerned. In addition, should lower-strength weld metals prove adequate in performance, weld metal deposition rates may be used that would provide possible cost saving.

The overall recommendation is that there should be a reconsideration of the weld-metal yield-strength requirement. Should undermatching (or even matching) strength be proven satisfactory, this lower strength level, in conjunction with improved low-hydrogen welding consumables and techniques, could lead to reduction or even elimination of hydrogen cracking; hence the need for preheating.

PREFACE

The Navy continuously strives to improve the efficient use of our natural resources and to reduce the cost of fabricating structures without sacrificing performance. A recent NMAB study for the Navy (Establishing Toughness Requirements for Low- and Intermediate-Strength Steels in Weapon Systems, NMAB-360, 1981), examined the status of fracture technology in the structural steels employed. Originally, that study was to examine the entire structure, but was narrowed to the base metal plates. The weldments added too many complex questions that needed separate examination.

Special handling considerations relating to cleanliness, quality control of materials and control of the welding process variables become a major part of the total fabrication cost. As the strength level of the steel is increased, it becomes more difficult to produce weldments of high integrity that possess properties equal to or better than those of the base plate. The question raised is whether or not it is possible and adequate to use, under controlled conditions, lower-strength weld metal that is more readily produced and applied. This lower strength weld metal would have the potential for higher toughness that might partially compensate for reduced yield strength.

The potential benefits in reduced fabrication cost and increased welding efficiency have been recognized. This study was directed to assess the effective use of the weld metal in naval structures and the rationale for necessary yield strength criteria.

NMAB assembled a group of experts recognized in the welding field to examine the feasibility of using lower yield strength steel weld metals, as much as 0.8 base metal yield strength of HY-80 to HY-130 streels, rather than the presently used matching and overmatching yield strength weld metals. This investigation focused on the areas of yield strength and toughness. The rationale for the establishment of yield strength and toughness criteria for the high-yield strength steel weld metals was reassessed and the need for an approach for generating rational weld yield strength and toughness requirements as addressed.

The committee in the course of its study determined, however, that more simplified fabrication procedures for HY-steels should be possible. These procedures involve reducing the sensitivity of weld to hydrogen-induced cracking (HIC) and minimizing the preheat and postweld inspection requirements. Recommendations are made on how this could be accomplished.

To cover all pertinent aspects of the examination, candidates who were considered for this study had to have expertise in the following fields: weld filler materials, welding technology, fracture mechanics of materials and weldments, and the economics of welding. Because of the multidisciplinary nature of welding technology, the committee has assembled in this document, information that is widespread in the literature. This resulted in a rather lengthy report whose broad coverage helps the reader grasp the nature of the problem directly without extensive library research. In particular, the Japanese contributions are not readily available elsewhere.

ACKNOWLEDGMENTS

The committee acknowledges with thanks the sponsoring agency for its support of this study program. The close cooperation received from the government liaison representatives was invaluable to the committee. They supplied considerable background information from their files that was helpful to the committee in focusing on various phases of the problem under investigation. In particular, P. Patriarca of the Breeder Reactor Materials Program, Oak Ridge National Laboratory, is thanked for his efforts on behalf of the committee in contacting various members of the American Welding Society for specific information.

The chairman wishes to thank J. D. G. Sumpler, the Admiralty Marine Technology Establishment (Scotland), for furnishing test data on weld metal yield strength (Appendix B). Committee member M. Masubuchi's associates at M. I. T., J. Agapakis and V. P. Papazoglou, took considerable responsibility in compiling with Dr. Masubuchi the extensive Japanese data reported in Chapter 8 and Appendix C. In addition, Dr. V. Papazoglou made an informal presentation before the committee of the pertinent results of his MIT doctoral thesis. These important contributions are acknowledged with thanks.

Special thanks go to I. Fiorite, P. M. Palermo and R. F. Swann of NAVSEA for their guidance in the early stages of the committee's deliberations and for their taking time to meet with the committee to discuss issues of concern. Finally, Dr. H. Kihara, President of the Japan Welding Engineering Society, is thanked for giving the committee permission to use many figures and tables from their extensive publications on the study topic.

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CHAPTER 1

SUMMARY OF CONCLUSIONS AND RECOMMENDATIONS

This review and discussion of existing welding technology and concepts led the committee to a number of conclusions, among which are the following:

1. Structures are designed on the basis of the yield or ultimate strength of the base metal and there is very little reliable evidence to support the requirement for overmatching a high-strength weldment system to achieve adequate performance, or, conversely, to show that slight undermatching will not provide adequate performance.

2. Economic advantages that would be made possible by using lower yield-strength weldments are a lower sensitivity to hydrogen-induced cracking; a reduction of preheat requirements, and the use of higher deposition rate processes; and the possible use of a single type electrode.

3. Consumables (welding electrodes) can be specifically formulated to give reliable strength, particularly for HY-80 and HY-100 weldment systems. For HY-130 weldment systems, a lower yield-strength requirement would facilitate the qualification of covered and low-hydrogen electrodes.

4. The predetermined cooling rate determines the welding parameters such as preheat, interpass temperature, and heat input; the higher the strength required, the more critically controlled must be the welding variables.

5. There is belief that some localized residual stresses can be as high as the yield strengths, even for HY-130 steels; the higher the yield strength, the higher is the potential residual stress and, hence, the sensitivity to hydrogen.

To use weld-metal yield strength more effectively, the committee makes the following recommendations for work on problem areas needing attention. No priority sequence was established for the recommendations because the committee feels that all have importance in the overall evaluation of the problem and that several could be lumped together in a well-planned research program.

1. Establish for weldments in HY-steel system realistic criteria for toughness based on service requirements and independent of welding process.

2. Verify experimentally that matching and undermatching weldments can meet performance requirements and determine the limit of undermatching at various strengths for different joint geometries. (The tensile, dynamic tear, and explosion bulge tests are applicable.)

3. Evaluate the hydrogen-cracking sensitivity of various HY steels with matching and undermatching weld-metal strengths as a function of welding process, hydrogen potential, and preheat level.

4. Determine the effect of preheating temperature on the magnitude and distribution of residual stresses for different levels of weld-metal yield strength.

5. Determine the effect of the preheating temperature on the critical hydrogen level that will cause cracking.

6. Determine the effect of weld-metal yield strength on the magnitude and distribution of residual stresses in weldments in base metals of different strengths.

7. Determine the hydrogen potential of available filler metals and consumables.

8. Examine the behavior of hydrogen in high-strength steel weldments to determine the distribution of hydrogen and residual stresses as a function of time.

9. Investigate methods for reducing the concentration of diffusible hydrogen in steel weldments.

CHAPTER 2

INTRODUCTION

The traditional approach to the design of high performance welded steel structures, especially in shipbuilding, has called for weld metal tensile strength at least equal to the specified minimum tensile properties of the base metal. Weld metal matching or overmatching of the ultimate or yield strength of the base metal is most generally required by code or appropriate fabrication document. On the other hand, toughness of the weld metal or of the base metal heat-affected zone (HAZ) is often less than that of the base metal, particularly in systems where the base metal is relatively strong. The toughness, if required by code, is often determined by the toughness that can be achieved reliably by good welding practice. The philosophy of matching strength combined with "adequate" toughness has evolved to ensure that the weldment was not a weak link in a structure. This is particularly significant in that the weld region is the most likely location for stress concentrations caused by such factors as fabrication defects, metallurgical discontinuities, and geometrical effects.

In recent years, with the development and increased use of high strength, quenched and tempered steels (HY-steels), extrapolation of the philosophy of matching or overmatching, with "adequate" toughness has been questioned. This is because, within a particular alloy system, as the mechanical strength increases, it becomes more difficult to maintain both strength and toughness in a weldment. In some other welding systems, it is impossible to achieve matching strength and adequate toughness. Many aluminum alloy weldments are made with lower strength weld metals; otherwise a low strength, heat-affected zone results from the welding thermal cycle. Nine percent nickel steel for cryogenic applications relies on fabrication procedures that permit the use of lower-strength, nickel-alloy filler metals.

This study was initiated in response to a request from the Department of Defense that the basic philosophy of overmatching the properties of high-strength steel weldments be reexamined. The reason for the request is to ascertain if satisfactory performance is achievable with matching or some degree of undermatching. Three major benefits could result: fabrication costs could be reduced, flexibility of design could be increased, and more latitude could be allowed in requirements for welding procedures. A side benefit would be a reduction or elimination of preheat requirements when using lower-strength weld metals for welding lower-strength steels to HY-steels.

Objectives

The NMAB Committee on Effective Utilization of Weld Metal Yield Strength was appointed to examine the feasibility of using weld metals of lower yield strengths than are currently required for high-strength steels. Consideration was to be given to weld metals with yield strengths as low as 0.8 base-metal yield strength for the high-yield-strength steels (HY-steels) in the HY-80 to HY-130 range.

Approach

Prime consideration was to be directed at yield strength and toughness, with the possibility that a trade-off between the two properties would lead to less costly fabrication procedures. Therefore, the first objective of the committee was to determine if major savings in fabrication costs could be achieved in this way. However, further analysis showed that it would be more important economically, to eliminate the preheating requirements for welding the HY steels. Thus the committee refocused its attention on the technical implications of eliminating preheat.

The effect of using weld metals of lower yield strength, combined with improved, low-hydrogen welding consumables (electrodes) available today, was examined as a method of avoiding weld cracking when using no preheat, at least for some noncritical weldments. The committee also examined the potential trade-offs of introducing weld metals into structures that match or even undermatch the strength of the base metals. Particular emphasis was placed on butt and fillet welds and their performance under plastic overload at high strain rates. Also of considerable interest to the study, however, are the requirements of less critical butt and fillet welds where HY-steels are joined to lower-strength HY-steels or other materials.

It was acknowledged that other considerations, such as fatigue and stress corrosion, cannot be ignored when reviewing changes in acceptance criteria. However, as a first step this study focused on the interrelated effects of yield strength and toughness on weldment performance. The committee divided the study into areas where committee members had particular expertise. The findings reported, however, represent the opinion of the committee as a whole. Each area is explored separately as it impacts on the problem and is covered in some detail to aid the reader to understand the various factors involved.

The report is divided into the following sections:

Industrial Considerations: Relative costs were evaluated to show where the benefits of using lower yield strength weld metal will be potentially most beneficial in the fabrication of HY-steels. Consideration is given to process applicability, welding-heat input, costs of filler metal, and procedural limitations relating to operations such as preheating and postweld heating.

Present Criteria for HY Steel Weldments: Data are presented on base metal and weldment properties required for HY-steel weldment systems. Methods of evaluation, typical spread of data, and the effects of welding variables are reviewed as they relate to yield strength and toughness.

General Review of Fabrication Codes: Military, and industrial codes, were studied for precedents and experiences relating to the use of undermatching weld-metal strengths and the rationale for acceptance.

Weld Metal Properties: Commercially available filler metals for welding high-strength steels are reviewed. Weld-metal composition is related to the welding process, welding variables, and deposition rates in terms of of the strength-to-toughness ratios of weld metals.

Hydrogen Cracking Tendencies of HY-Steels: The susceptibility of HY steels to hydrogen-induced cracking (HIC) is reviewed both for weld-metal and heat-affected zones. The influences of the welding process, welding procedures, and strengths are considered.

Strength of Weldments in High Strength Steels: Literature is reviewed on undermatched weldment systems where both experimental and theoretical predictions of weldment performance have been made. Effects such as weld metal reinforcement, triaxial restraint in thick weldments, and residual welding stresses are considered.

CHAPTER 3

INDUSTRIAL CONSIDERATIONS

In the building of ships and submarines using HY-80 and HY-100 steels, the cost of welding is the shipyard's single largest direct labor factor. Welding, together with other attendant costs, is as high as 20 percent of the total shipyard costs. A typical welding department in a shipyard employs approximately 10 percent of the total labor; nearly double that of any other single trade. Therefore, welding and associated trades represent an area where a small percentage of improvement represents a significant overall cost savings in shipbuilding.

The major factors that contribute to the overall cost of welding are the metal deposition rate (lb/hr.), the welder operator factor (arc-hours/total hours), inspection, consumables (filler metal, shielding gases, flux, etc.), and in the case of HY-steels an additional cost for preheat must be included. The preheating cost includes the installation of preheating devices, the cost of energy, and additional labor costs resulting from a lower welder operating factor, and a general disruption factor that involves workers other than the welding trades. The preheat and the associate costs, therefore, significantly increase the cost of welding HY-steels over that of conventional construction steels. If preheat were to be eliminated or reduced, significant cost savings could be realized. It has been estimated that the elimination of preheat for certain attachments (non HY-steels) to HY-steels, as well as welding HY-steels to itself in thicknesses of 1/2-inch and under, can provide savings in excess of one million dollars per ship (1981 dollars).

Welding cost (i.e., dollars per lb of deposited weld metal) can be explained by reference to Figure 3-1, which shows the interrelation of welding process used, the weld-metal deposit rate, and the operator factor, to total cost of the weld metal deposited. Welding costs specific to HY-steels are shown in Figure 3-2. The labor costs, in both figures, for these analyses are assumed at $15 per hour and do not represent actual shipyard costs. The current fabrication requirement for the use of HY-steels in submarine construction, incorporated in NAVSHIPS 0900-006-9010, requires the application of preheat and specifies the applicable processes and filler metals. The filler metal for the shielded metal arc (SMA) process requires the use of electrodes that have yield strengths as high as 98 ksi. Other higher-deposition processes, such as the flux core process, are restricted partially because the higher yield strength requirements cannot be easily achieved and still meet the toughness requirements.

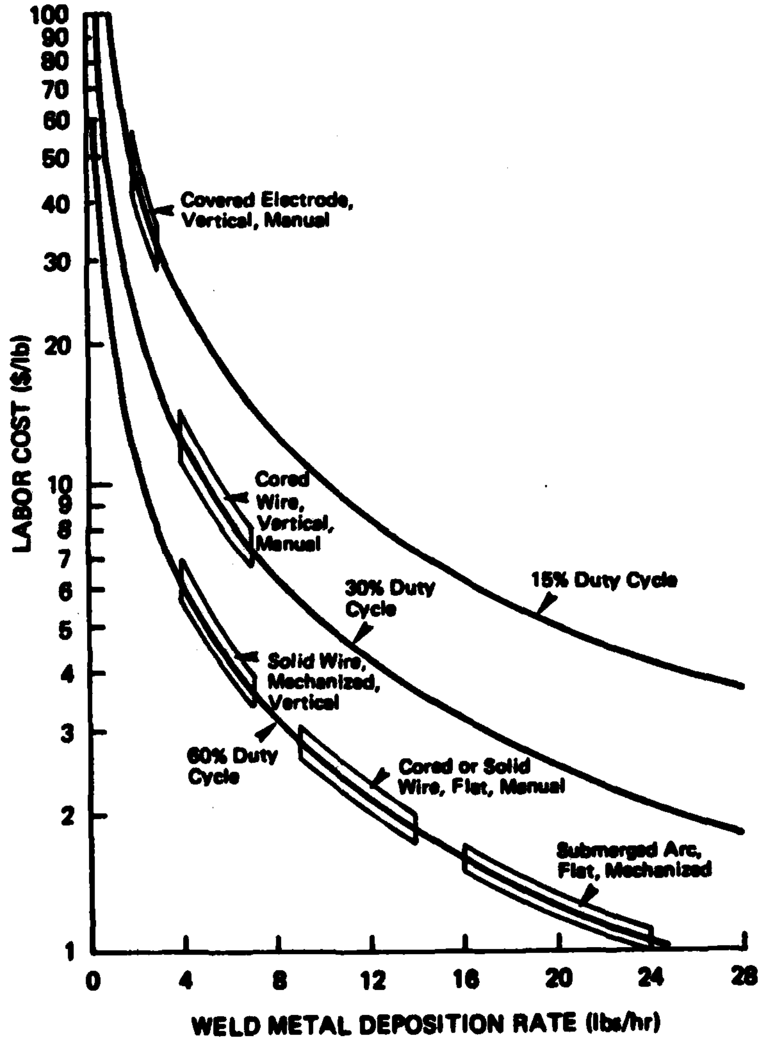

Figure 3-1 Effect of process and labor efficiency on test of welding-related costs for weld metal.

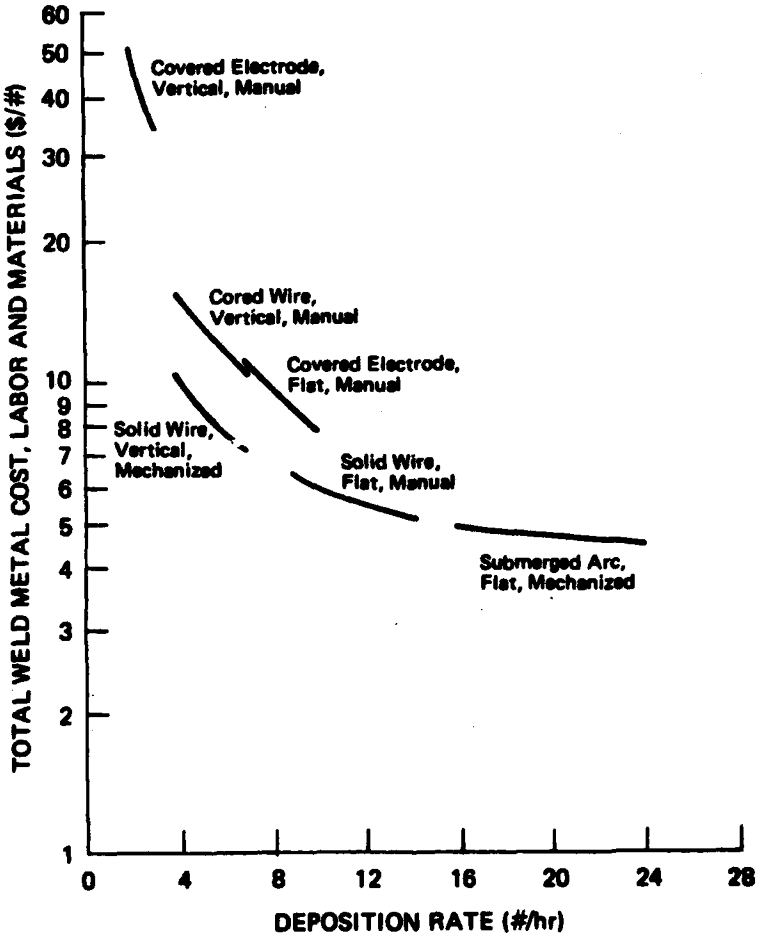

Figure 3-2 Relative costs of welds in HY-80 as affected by welding processes.

A major cost factor arises from the requirement for preheat that affects the cost in several ways. One is the high cost of the energy needed to achieve the preheat elevated temperature requirement. Another and equally significant cost is the reduction in the operator factor associated with preheat. Operator factors as low as five percent can be expected where the application of preheat is required. This factor does not include disruptions caused by other trades during the pre-heating period. It can be concluded that significant construction cost savings can be achieved by the elimination of preheat. The existing specifications for welded fabrication of HY-80 requires the use of overmatching yield-strength filler materials. Although the permitted yield strengths vary with the processes used, the greatest specified overmatch is that for manual shielded metal arc welding (SMAW) where the yield stength overmatch is approximately 120 percent of the minimum base-metal yield strength. The SMA process represents more than 70 percent of all welding processes used. In addition, magnetic particle testing (MT) inspection is required to assure that weld toe or surface cracks have not developed. For certain applications, MT inspection is required seven days after final welding because of delayed cracking. Controls are also placed on the initial moisture content of the electrode that is to be exposed to the ambient atmosphere. Similar requirements are also imposed on welds for attaching lower yield-strength materials to HY-80. These requirements for preheat and interpass temperature controls, the post-weld MT inspections, and electrode controls result in increased fabrication costs for HY-80 steels.

The requirements for preheat and interpass temperature controls together with electrode moisture control are imposed principally to avoid delayed hydrogen-induced cracking. To significantly reduce fabrication costs, however, the costs related to preheating must be minimized while at the same time ensuring crack-free weldments. To safely reduce costs, it is necessary, therefore, to evaluate the factors that contribute to hydrogen-induced cracking and the means to minimize their effects in a cost-effective way. One cost-effective technique would be to modify preheat requirements by assessing the potential advantages of using lower yield-strength weld metal coupled with low-hydrogen processes.

The factors and interrelationships that cause hydrogen-induced cracking when welding HY-80 steels are dealt with in more detail in later chapters. Briefly, however, they are (1) the presence of hydrogen, (2) a sensitive microstructure, and (3) the stresses across the weld zone resulting from cooling of the weldments and imposed loads. In a given HY-80 weldment system, the alloy content (microstructure) necessary for hydrogen cracking is present. The controllable factors, therefore, are the level of diffusible hydrogen and the restraint stresses across the weld zone due to the cooling weldment. The latter is a function of the yield strength of the filler metals.

Current fabrication requirements reduce the effect of hydrogen by controlling electrode moisture and by imposing preheating. However, the cost of preheating can be eliminated provided it can be substantiated that weld-related stresses are reduced by using lower yield-strength electrodes in combination with lower-hydrogen processes. Thick plate weldments and

associated high-restraint levels may necessitate preheat even with lower yield-strength weld metals. Nevertheless, economic advantage may result from the elimination of preheat and possibly postweld nondestructive testing (NDT) requirements when joining lower yield-strength materials (such as attachments) to HY-80 using lower yield-strength filler metal.

A further economic consideration favoring the use of lower yield-strength weld metal is the possibility of introducing more economical high-deposition processes, such as the flux-core process, also using a single type of electrode when welding lower yield-strength material to HY-80.

Restraint Cracking

Restraint Stress and Cracking

As previously stated, the major measures that make HY-80 more costly to fabricate than other construction steels are those necessary to control hydrogen-induced cracking. If it can be shown that for a given level of elemental or diffusible hydrogen, cracking becomes simply a function of the applied and residual stresses in the system. A reduction in these stresses by the use of lower yield-strength filler metal should reduce the cracking tendency.

Preheat, Stress and Cracking

Should studies indicate that there is a direct relationship between hydrogen-induced cold cracking, levels of preheat, and stresses, either residual or applied, then, for a given reduction in stress, a corresponding reduction in preheat should be tolerated. Extending this relationship, it should then be possible to eliminate preheat for certain combinations and thicknesses of materials and for certain joint designs. Specifically, it should be possible to eliminate preheat when attaching lower yield-strength materials to HY-80 using low-strength electrodes.

Cracking, Electrode Moisture and Stresses

If studies indicate that cracking is a function of absorbed hydrogen and weld-metal yield strength, as it is believed to be, then, for a given preheat temperature and level of hydrogen, the degree of cracking will vary with the level of induced or residual stress--that is, as the stress is reduced, cracking is also expected to be reduced. Similarly, as the electrode moisture or diffusible hydrogen is reduced, cracking can also be expected to be reduced or eliminated. It is therefore important to consider lower available process moisture in combination with reduced preheat and lower yield strength electrodes.

Reducing Preheat Requirements

As was previously stated the major added cost of fabricating HY-80 involves preheat, and a major cost reduction could be achieved by eliminating preheat when possible. One potential place to eliminate it would be where lower yield strength materials are used for joining attachments to HY-80. This is where electrodes of lower yield strength and lower moisture content could be used. This consideration is based on the reduced sensitivity to hydrogen cracking associated with lower yield-strength materials and filler metals and the use of lower-hydrogen processes.

Preheat and Preheat Control

Preheat Requirements

The existing preheat requirements for welding HY-80 vary with the thickness of the materials and with the welding process (Table 3-1). It is apparent that the effect of yield strength on levels of restraint has been recognized by permitting lower preheat for thinner sections. It should be pointed out, however, that the diffusion path for hydrogen is also reduced in thinner members, which contributes to the lower susceptibility to cracking. In addition, the lower-hydrogen processes, such as gas metal arc (GMAW) and submerged arc (SMAW) welding processes also have been recognized as lowering cracking susceptibility by permitting lower levels of preheat. Similarly, lower preheat temperatures are permitted for the lower yield-strength SMAW (MIL 9018) electrodes. Furthermore, the requirements for preheating are reduced for welds involving austenitic filler metals, and postweld nondestructive testing (NDT), such as magnetic particle testing (MT), is not needed.

TABLE 3-1 HY-80 Preheat Requirements

Thickness (inch)	Preheat/Interpass Minimum (°F)			Preheat/Interpass Maximum (°F)
	MIL-11018	MIL-9018 Gas Metal Arc Sub Arc	Austenitic* Electrodes	
1 1/8 and greater	200	150	125	300
From 1 1/8 to 1/2	125	150	125	300
1/2 and less	60	60	60	300

* Post weld NDT is not required when using austenitic electrodes.

Source: MIL-STD-1688 (SH), Fabrication,, Welding and Inspection of HY-80/100 Submarine Applications, February 9, 1981, NAVSEA, Washington, D.C.

Preheat and Hydrogen Cracking

The reason for preheating in HY-steels systems is to reduce cracking. It is generally accepted that the effect of preheat is to extend the time the weld is at elevated temperature, thus permitting greater quantities of diffusible hydrogen to dissipate from the weld zone (see Chapter 6). This is unlike the case of some high alloys, where the primary purpose of preheating is to affect the microstructure; in HY-steels it serves to promote slower cooling thus extending the time for the hydrogen removal. It should be recognized, however, that in fabricating any large constrained structure, preheat has the added advantage of promoting more uniform cooling and so reducing cooling stresses and the likelihood of cracking.

Cost of Preheat

The cost of preheating involves many direct and some indirect factors. These are identified and discussed later. It should be emphasized, however, that the cost of preheating is the single most significant contributor to the higher cost of fabricating HY-80 compared with other constructional steels. Therefore, the elimination of preheat would do most to reduce fabrication costs. This is true whether it is eliminated for certain thicknesses or just for simple attachment welds. Even a reduction in preheat temperature would reduce costs. However, the savings would be minimal and essentially would relate only to reduced power consumption, not to the more significant cost factors.

The main cost factors associated with preheating are:

<u>Facilities</u>: Requirements that include the central power station, switch gear, and distribution capacity.

<u>Energy</u>: In a shipyard, the cost of energy for preheating far exceeds the cost of all other energy required for welding.

<u>Preheating Devices</u>: These costs include the capital cost of the devices as well as their replacement costs. Most preheating devices have relatively short lifespans.

<u>Application of Devices</u>: The direct labor involved in attaching the heaters represents a sizable cost factor.

<u>Trades Disruption</u>: After fitting and assembly, work must be planned and executed around preheaters. Other trades have limited access to work areas when preheaters are energized.

<u>Reduced Productivity</u>: Workers' productivity suffers from working on and in preheated (200°F-300°F) structures, particularly during warm weather.

Assembly and Outfitting Delays: The requirement for preheating when making attachments to HY-80 requires preheat during advanced stages of construction. There is waiting time for the proper temperature to be achieved and other activities can be disrupted.

The requirement for preheating and controlling the interpass temperature results in the most significant costs associated with the fabrication of HY-80, and by far the most significant of these is the cost of applying preheat devices and maintaining their effectiveness. One can visualize large butts and seams equipped with preheaters and located in open spaces and conclude that preheating costs may not be too significant. Less apparent, however, are the thousands of applications of preheat to attachments, brackets, and foundations, all within a ship's hull and many involving attachment of lower yield-strength materials to HY-80. The cost of applying preheat to a small bracket or hanger may well equal or exceed the cost of fitting and welding.

Electrode Control

Electrode Exposure Time

Currently, because of the hygroscopic nature of electrode coverings and the desire to minimize hydrogen damage, five hours is the maximum permitted time of exposure of a covered electrode to the ambient atmosphere. This requirement results in electrodes being issued at least twice during a normal eight hour shift. This practice disrupts work and reduces productivity. This disruption could be eliminated by using a lower yield filler metal and more moisture-resistant electrodes and so extending the time electrodes may be exposed to the atmosphere.

In recent years, considerable progress has been made in producing moisture resistant electrodes. Curves can be generated (chapter 5) that show that electrode coverings can be produced that have low initial moisture and maintain low moisture for exposure periods exceeding the normal eight hour shift under rather extreme conditions of temperature and humidity. The advantages of these developments in conjunction with lower yield-strength weld metals should be recognized when considering modifications to preheating requirements.

Baking, Storage and Handling of Electrodes

Current practice has been to rebake electrodes after receipt from the manufacturer to ensure the proper moisture level during welding. However, recent changes in these requirements reflect the advantages of the new lower-moisture electrodes.

Initial Baking Requirements: The cost of rebaking involves handling and the restocking of electrode-dispensing stations as well as the baking process itself. The additional handling often results in losses from breakage of electrode coatings, etc. Again, of course, facilities and power are required for this operation.

Use of a Single Type of Electrode

The advantage of using a single type of electrode for joining all steels to HY-80 may not be readily apparent and in fact may seem inconsistent with the objectives of this study. However, many structures in which HY-steels are used include grades of steel other than HY-grades. Usually these are high-tensile (HT) steel having a yield strength of about 50 ksi and mild steel having a yield strength of about 35 ksi. The problem sometimes encountered is the identification of material grade during fabrication. Identification is particularly troublesome in enclosed areas and tanks where the lighting may be poor. It would thus be desireable to use a single type of electrode that is compatible with all grades of steel to be used.

The anomaly is that, when arguments are made to reduce the yield strength of weld metal for HY-steels, one must then justify the use of higher yield-strength weld metal for joining lower-yield material. The justification is that the concern with overmatching the HY-system involves a microstructure that is sensitive to hydrogen cracking, and in welding the lower-strength materials (50 ksi yield and lower) that do not have the sensitive microstructure, this concern does not exist. It is therefore desirable from a construction viewpoint to issue a single electrode, similar to the E9018, to satisfy the overmatching problem in HY-system without deleteriously affecting the lower-strength grades.

Discussion

The technical and economic factors discussed in this chapter indicate that if it is technically feasible to reduce or eliminate preheat in welding materials in the HY-system by the use of lower yield-strength filler metals and those of lower diffusible hydrogen, significant cost savings can be realized. It is also possible that in welding attachments to HY-steels--particularly attachments made from materials with lower yield-strengths (less than 50 ksi)--many of the preheat and postweld inspection requirements modification or elimination may be achieved and further reduce total costs.

Table 3-2 is a matrix relating preheat and inspection requirements for joining HY-80 to itself and for joining lower yield-strength materials to HY-80 in various ranges of thickness. The thickness ranges selected are those that are used in the current fabrication documents and do not necessarily reflect the ultimate ranges where preheat and inspection requirements may be modified. Table 3-2 is intended only as a guide to pinpoint areas where it may be currently feasible to reduce or eliminate preheat and the postweld NDT

TABLE 3-2 Preheat and Interpass Temperature and Magnetic Particle Testing (MT) Requirements (for Welding HY-80, and HY-80 to 50 ksi and less steels)

Base Material Grade to Grade (a)	(b)	Thickness, in. Grade to Grade (a)	(b)	Filler Metal Yield Strength ksi min.	Preheat/Interpass Temp. °F min.	Postweld Mag. Part. Insp.
HY-80	HY-80	1-1/8 & over	1-1/8 & over	82	TBD	Required
		1-1/8 to 1/2	1-1/8 to 1/2	82	TBD	TBD
		1/2 & less	1/2 & less	82	60	TBD
HY-80	HY-80	1-1/8 to 1/2	1-1/8 & over	82	TBD	Required
			1/2 & less	82	TBD	TBD
HY-80	HY-80	1/2 & less	1-1/8 & over	82	TBD	TBD
			1-1/8 to 1/2	82	TBD	TBD
HY-80	50 ksi YS & less	1-1/8 & over	1-1/8 & over	55	TBD	TBD
			1-1/8 to 1/2	55	TBD	TBD
			1/2 & less	55	60	Not Req'd.
HY-80	50 ksi YS & less	1-1/8 to 1/2	1-1/8 & over	55	TBD	TBD
			1-1/8 to 1/2	55	TBD	TBD
			1/2 & less	55	60	Not Req'd.
HY-80	50 ksi YS & less	1/2 & less	1-1/8 & over	55	60	Not Req'd.
			1-1/8 to 1/2	55	60	Not Req'd.
			1/2 & less	55	60	Not Req'd.

Note: 1. Low moisture electrode, i.e., 0.1 percent initial moisture with controlled absorption rates.

2. Thickness range based on existing specified range. May change after testing.

TBD = to be determined; test required to establish minimum material thickness; range may also be modified.

inspections. It also points out areas where investigation is required to adequately establish new limits of preheat and NDT procedures for various ranges of thickness and combinations of materials. Similar matrices can be developed for other strength levels in the HY-steels.

The filler metals considered in Table 3-2 are assumed to be those with low initial moisture (i.e., 0.1 percent or less) and with demonstrated low moisture-absorption rates. The deposited yield strength selected for joining HY-80 to itself is 82 ksi; the minimum for filler metal joining HY-80 to lower yield-strength material (50 ksi and less) is 55 ksi.

In most cases, the minimum preheat temperatures are indicated as TBD (to be determined), but for certain conditions, technical support is currently adequate to establish new, modified values. For these welds, residual stress data is needed to support the lower preheat levels associated with the lower residual stresses expected to result from using low yield-strength weld metal and thinner and lower-strength materials.

Consideration is given to the elimination of magnetic-particle testing (MT) where the probability of hydrogen cracking is near zero--for example, when welding light attachments using austenitic electrodes. The objectives of any future studies should include the reduction or elimination of postweld MT inspection, since it is a significant economic factor.

Conclusions

1. It would be desirable to use matching or near-matching yield-strength in the fabrication of HY-steels and thus all welds, including attachment welds, to be made with a single type of electrode and with a reduced need for or the elimination of preheat.

2. Modifying the requirements for preheat and interpass temperature could permit the use of matching yield-strength metal in combination with lower-moisture covered electrodes.

3. Using undermatching electrodes with lower-moisture covered electrodes could reduce or eliminate preheat and interpass temperature requirements where appropriate.

Recommendations

The following actions should be taken to increase productivity and reduce cost without sacrificing structural reliability:

1. Eliminate or modify requirements for preheat and interpass temperature where technical data justifies such action.

2. Eliminate or modify magnetic particle inspection (MT) where technical data indicates that hydrogen cracking is not of concern.

3. Conduct test programs in areas where more corroborative data are needed to show that preheat and postweld MT inspection can be eliminated.

CHAPTER 4

MATERIALS AND FABRICATION REQUIREMENTS

Explosion bulge tests by the U.S. Navy in the early 1950s, using photogrid for the weld metal tested in biaxial stress fields, demonstrated that overmatching caused "relief." For conditions of plastic deformation when the action of stresses were across the weld, the photogrid measurements demonstrated that low strain developed in the weld compared to the base metal and that this effect was proportional to the degree of overmatching. The concept of protecting the weld metal by overmatching originated from this work (Pellini and Hartbower 1951).

For the HY-80 system, explosion bulge tests, during the period 1954-57, demonstrated that similar "relief" with associated low strain developed in the weld compared to the base metal when overmatching was used. These tests also showed the effects of weld geometry; single Vee, double Vee and associated groove angles. By adjusting of weld groove angles and with weld metal overmatching, it appeared that fissuring or cracking of the weld and HAZ could be suppressed--i.e., the weld zone could be protected from strain localization effects. The thesis that overmatching for HY-80 is desirable developed largely from this very general experience. Conversely, no information was available to show undesirable effects of overmatching.

There is a general consensus that shielded metal arc welding (SMAW) with stick electrodes produces welds with less homogeneity than automatic SMAW. In addition, surface contour is generally less favorable for the SMAW than for the SAW process. Since overmatching was easy to obtain for HY-80 when using SMAW with stick electrodes, overmatching became the general practice for SMAW. Later, about 1956, the first electrode for use with HY-80 was MIL-26015(16); a low-hydrogen ferrite electrode. It had more than adequate strength, but its Charpy V-notch toughness was only 20 ft-lb at 0°F. This electrode was the early version of the currently used E11018M electrode.

The concept of overmatching of filler metals for automatic SAW, and the later manual semiautomatic gas metal arc welding (GMAW) for HY-80, was not continued. Service experience with all these processes has been satisfactory and should substantiate or indicate that overmatching is not a necessity for SMAW or any other welding process.

Strength and Toughness--HY-80/-100/-130 Systems

Experience with failure in bridges, pipelinès, aircraft, rocket cases and ship hulls provided sufficient evidence that simple strength is no more important and may be less important than the other properties of the materials. For steels with yield-strengths above approximately 80,000 psi (80 ksi) transition from high- to low-fracture tolerance is based primarily on strength and quality. In general, as the temperature decreases or the yield-strength or rate of loading increases, the tolerance to flaws decreases. In the presence of a flaw, a tensile-stress field, and proper temperature conditions steels can fracture in one of three modes often characterized as:

Brittle: Low energy fracture, considerably below the theoretical limit, without significant deformation under elastic stresses.

Elastic/Plastic: Mixed mode fracture with considerable deformation, partially reversible, requiring higher normal applied stresses and larger flaws for the onset of the strain and fracture.

Ductile: Plastic deformation followed by strain hardening before brittle fracture occurs.

It is most desirable to utilize steels that will fracture in a ductile mode, but with proper engineering analysis, high elastic-plastic materials can be used *safely (Palermo* 1976).

A minimum level of toughness has been selected for the HY-steel submersible weldment systems to preclude the catastrophic propagation of a crack at any elastic stress level for an assumed design temperature limitation of 32°F (0°C). In addition, explosion bulge testing is carried out to assure ductile performance at 0°F for HY-80 and HY-100 steels and at 30°F for HY-130. These measures provide for structural reliability under explosive overload. The specified criteria for toughness, the controls imposed on welding parameters, and nondestructive inspection procedures used for naval structures assure this high level of performance.

Plastic performance even in the presence of full-thickness flaws is obtained by a metallurgical balance of composition and heat treatment. This balance ensures that the brittle-fracture transition region is below the minimum temperatures of 0°F for HY-80 and HY-100 and 30°F for HY-130 products. This performance level is considered to be a "yield criterion" performance level that is described in terms of the linear elastic fracture mechanics (LEFM) parameter β , and when β equals one, small cracks are stable and pop-in cracks are arrested. The equation for β is:

$$\beta = \frac{1}{B} \frac{K_{Ic}}{\sigma_{ys}}$$

where β = the beta parameter

B = thickness

K_{Ic} = critical plane-strain stress-intensity factor

σ_{ys} = yield-strength

If conditions exist such that β equals one, then local, through-thickness yielding will occur allowing surface flaws to penetrate the thickness without becoming unstable. Thus, any flaw of less than full thickness is stable at all elastic stress levels and small pop-in flaws from local brittle regions will be arrested before achieving full thickness.

The criteria for fracture toughness in the material specifications are in terms of small scale test results, but the criteria are based on analytical concepts. The charpy V-notch (C_v) or dynamic tear (DT) test criteria for base materials assure fully ductile, plastic performance of sections of more than 2 in. thick at 0°F for HY-80 and HY-100 products and at 30°F for HY-130 products. The C_v or DT test criteria for the weld metals assure a toughness sufficiently high so that any detectable flaw (possibly a flow of at least 1/4 of section thickness) will require gross plastic overloading (greater than 2 percent plastic strain) to penetrate the section. This performance must be demonstrated in the crack-starter explosion bulge test. Although welds are prone to have small fabrication flaws, the depths of undetected flaws are substantially less than 1/4 section thickness (because of in-process supervision and nondestructive testing after welding) and these small flaws are stable even in the presence of plastic strains in excess of 5 percent as proven and certified by the explosion bulge test requirements for weldment performance. Therefore, small reductions in the minimum yield-strength of weld metals would have an inconsequential effect on the extension of any such flaw since 10 ksi stress is equivalent to less than 1 percent of the total strain that is involved in the explosion bulge tests.

Fracture toughness criteria used in the specifications for weld metals are based on an empirical relationship between DT energy and K_{Id} (Figure 4-1). The appropriate DT energy for a specific product is calculated from toughness (K_{Ic}/σ_{ys}), section thickness (B), and the relative structural performance factor, (β). Most of the DT energies associated with $\beta = 1.0$ performance are in the temperature-transition region for the HY steels. For this reason, the effect of particular section thickness and related constraint conditions on toughness must be considered when extrapolating small-scale test data to full thickness performance for establishing an acceptance criterion. To predict the performance of the thicker section accurately, a correction must be applied to the temperature at which a 5/8-in. section reaches the desired level of toughness (K_{Id}). A temperature shift of 50°F is needed to correct for the effect of the added constraint in a 2-in. thick section. An example of how this correction is applied to 5/8 (0.625) in. DT energy values is illustrated in Figure 4-2. The 50°F temperature shift shown in Figure 4-2 is based on the results of a series of Naval Research Laboratory (NRL) studies on the effects of constraint on the transition temperature curve for sections up to 12-in. thick. It was found that the mid-transition temperature of the 5/8-in. DT energy curves for all of the steels tested were consistently 50°F lower than the mid-transition temperature of the 2-in. DT energy curves.

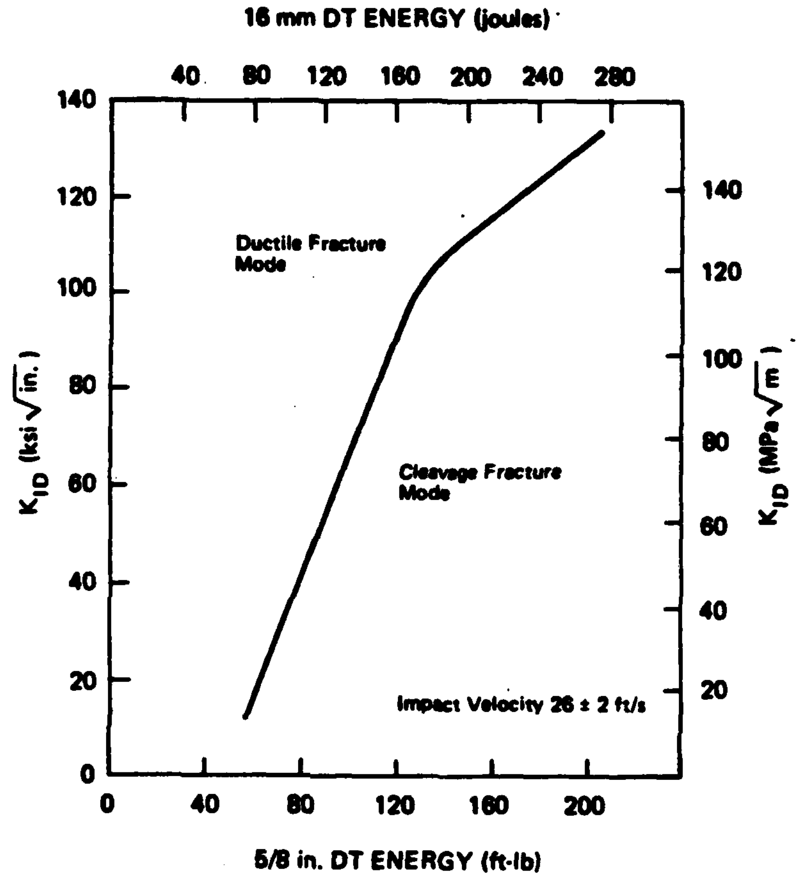

Figure 4-1 Empirical relationship between DT energy and toughness.

Source: Lange 1977.

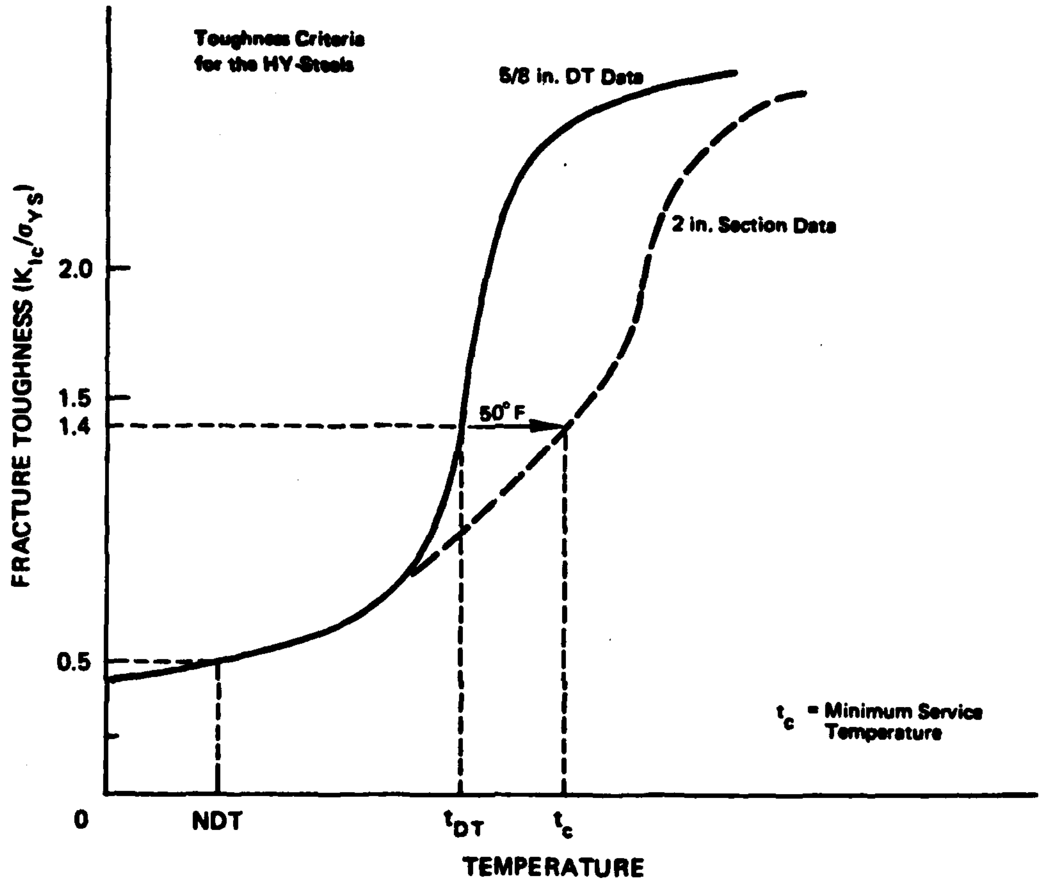

Figure 4-2 Illustration on the selection of DT test criteria for the HY-steels.

Source: Lange 1977.

Plastic performance is a three-dimensional phenomenon so the two-dimensional analysis commonly used in linear elastic fracture mechanics (LEFM) is not appropriate. When plastic strains are required to propagate cracks, small changes in yield strength, such as 10 ksi, which represents a plastic strain of approximately 0.03 percent, have an insignificant effect on the fracture performance of a weldment. On the other hand, if the weld metal yield strength is not exceeded by normal loud stresses, the weld performance is not effected even for the case of an undermatching condition. This is because the basic LEFM equation that relates stress intensity (K_I) to applied stress (σ) and flaw size (a) does not include a yield strength parameter:

$$K_{I\ (appl)} \sim \sigma_{(nom)} \sqrt{a}$$

Toughness values need to be worked out for each HY-80, HY-100, and HY-130 weld-metal system to ensure adequate service performance (fitness for purpose). These values should relate to a specific service temperature; for underwater ocean service a minimum of 0°C (32°F) is generally assumed. Using data from other small- and large-scale toughness tests, toughness values can be determined for each weld-metal system. These toughness levels will change with stress level. Since the minimum yield strength of the E-9018-M electrode is 78 ksi, a full range of stress from 78 ksi to 130 ksi should be considered for HY-steels.

Small-Scale Tests

Charpy V-notch tests today form the basic small-scale test requirement for HY-80, HY-100, and HY-130 systems. In some specifications provisions are made for dynamic-tear tests. Drop weight tests also are conducted to evaluate nil-ductility temperature. Toughness data should be examined for each welding process and it is recommended that a single test temperature be established for each test method. A single energy requirement is also recommended regardless of weld process.

Explosion Bulge Tests

The methods, procedures, and acceptance criteria for explosion bulge tests are in NAVSHIPS 0900-005-5000, "Standard for Preproduction Testing Materials by the Explosion Bulge Test" (1965). [Revision and updates of NAVSHIPS 0900-005-5000 was proposed by the W. D. Taylor Naval Ship Engineering Center and commented on by P. P. Puzak in 1971; see Appendix D, W. S. Pellini's letter of March 2, 1971 with attachments.] The explosion bulge test was developed during the period of 1949-50 and has been used extensively to investigate the factors that determine the performance of weldments, particularly in submarines and other large welded structures. The test determines reliably the performance characteristics of service-type weldments and is the only feasible way to evaluate fully the heat-affected zone of weldments and the performance of weld metal.

Preproduction tests of base materials, filler metals and welding processes are performed in accordance with the NAVSHIPS standard cited above. When there is a major alteration in any part of the approved manufacturing or approved welding process, requalification is required of the product or process.

The explosion bulge test is expensive and can be done at a limited number of facilities. It is used to assess weldment performance and for testing materials. The base material, weld, heat-affected zone, and fusion line in weldments are evaluated in a single test. An American Bureau of Shipping (ABS) Report (March 1975) presents comments on the use of this test to evaluate electroslag and electrogas welds in ordinary and higher-strength ship steels. Small-scale tests are performed and evaluated prior to bulge testing. Crack-starter explosion tests generally are carried out for screening before explosion bulge testing. The purpose of the crack-starter tests is to produce an early crack that could result in catastropic propagation of fracture if the material, weld, heat-affected zone, or fusion line has a tendency toward low-energy crack propagation. Explosion bulge tests require the repeated explosive shots to delineate the regions of the weldment that are critical in the initiation and propagation of fracture.

The NAVSHIPS document states that, "Test temperatures shall be as specified by the Bureau of Ships." For HY-80, the test temperature generally used is 0°F, however, test temperature of 32°F, the lowest expected under water, should be considered. The acceptance criteria for the explosion bulge test are believed to be rather conservative. On this basis, materials or weldments that perform satisfactorily in the bulge test are judged acceptable for critical structures that are both statically and dynamically loaded.

The long-term satisfactory performance of HY-80 weld systems appears to indicate that the explosion bulge test should provide the background data for establishing the governing criteria for small-scale static tests. A program to evaluate matching and undermatching of weld-metal yield strength should be considered. Weld metals are available with matching or slightly undermatching (by about 10 to 20 percent) yield strength that exhibit various toughness properties as measured by small-scale tests. Correlations need to be established between yield strength and small-scale toughness based on explosion bulge criteria.

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NAVSHIPS 0900-005-5000 (formerly NavShips 250-637-6), Standard Procedures for Preproduction Testing Materials by the Explosion Bulge Test, Revision 1, November 1965, Bureau of Ships-Navy Department, Washington, D.C.

CHAPTER 5

REVIEW OF U.S. INDUSTRY WELDING CODES

The incorporation of various written welding procedures into a codified structure by identifiable end use has produced a variety of codes and standards. Several industrial and professional organizations have published welding codes specific to particular applications. The objective of this review is to determine if precedent exists in the various welding codes for undermatching weld-metal strength (as compared to base-metal strength).

ASME Boiler and Pressure Vessel Code

The ASME Boiler and Pressure Vessel Code (B&PV Code) is divided into two basic types of sections: those pertaining to the design, fabrication, inspection, and testing of specific components or systems; and those dealing with general material, process, or inspection requirements. Because of the special needs of each type of component however, the rules of the latter sections may be supplemented by the component section. For the purposes of this review, the two component sections that deal with pressure vessels and structures (Sections III Nuclear Power Plant Components and VIII Pressure Vessels), have been selected as being most closely related to this study. The requirements for weld-metal and welding procedure qualification were reviewed with respect to the question of undermatching strength and toughness.

Base and Weld Metal Requirements

The base metals used in B&PV Code components must be produced in accordance with specifications in Section II or in special Code Cases. Each component section of the code specifies which of these metals may be used and to what extent. For example, Section III allows SA-542, Class 2 (2-1/4 Cr-1 Mo) plate to be used for component supports, but not for pressure vessels. Although Section II also includes specifications for welding material, the use of filler metal not listed in this section is not necessarily prohibited, provided the material used is tested, identified, and controlled.

The B&PV Code includes steels similar in strength to HY-80. However, no materials with minimum yield strengths of 100 ksi or higher (HY-100 and HY-130) are currently allowed for welded construction.

Welding Procedure Qualification

Once the base metal(s) to be joined have been selected, the welding procedure must be proven. The basic qualification requirements are contained in Section IX, Welding and Brazing Requirements. This section (IX) specifies that a written procedure must be prepared that contains the welding parameters: the Weld Procedure Specification (WPS). To qualify a new WPS, coupons are welded and tested in accordance with Section IX and supplemented by the applicable component specification. Tests on tensile, bend and impact behavior provide assurance that the required strength, ductility and toughness will be found in the fabricated product.

In an attempt to minimize the number of qualifications an organization is required to perform, Section IX requires requalification only when changes are made on the weld parameters that affect mechanical properties or some other performance requirements. These parameters are termed essential variables (see Tables QW-252 through QW-261, Section IX, B&PV Code). Examples of essential variables are welding method, base metal, filler metal, base-metal thickness, and postweld heat treatment. Base metals having similar weldability characteristics have been grouped under "P" numbers and "P" number subgroups. A welding procedure qualified for a particular P number and group is usually qualified for any other material within the classification.

Quenched and tempered low-alloy steels similar mechanically and metallurgically to the HY-steels are grouped under P numbers 3 and 11. P-3, Group 3, includes those quenched and tempered materials of 80 to 90 ksi tensile strength that generally represent the majority of material used in nuclear pressure vessels (Section III). P-11A includes quenched and tempered material similar to HY-80 (i.e., minimum tensile strength of 100 ksi and minimum yield strength of 80 ksi). P-11B includes material with slightly higher minimum tensile and yield strengths (115 ksi and 85 ksi, respectively). These represent the highest-strength materials allowed for welded construction. Therefore, the rules for welding P-11 quenched and tempered low-alloy steels are the most applicable to the HY series.

Nuclear Power Plant Components

Section III of the B&PV Code, Nuclear Power Plant Components, is divided into Divisions 1 and 2, with Division 1 being further divided into six subsections covering various classes of components. Reference will be made to Subsection NB, Class 1 Components; these being typical of applications under consideration here.

Subarticle NB-2400, Welding and Brazing Material: Requirements for the testing of weld filler metal in Subarticle NB-2400 are consistent with the requirements for qualifying base materials. The types of tests, specimen location, preparation, and acceptance criteria reflect the language used for base material. In particular, weld metal must have the minimum tensile strength specified for the base material. When base materials of different specifications are welded, the tensile-strength requirement is the mimimum specified for either base material.

Fracture toughness testing for weld metal is required when impact testing is required for either base material of the production weld. The acceptance criteria for the weld metal conform to the requirements of either base material.

Subarticle NB-4300, Welding Qualifications: Subarticle NB-4330, "General Requirements for Welding Procedures Qualification Tests," details preparation, orientation, and testing requirements. This subarticle supplements Section IX. While other sections of the code cover test and acceptance requirements for base (NB-2800) and weld (NB-2400) materials, NB-4330 is the only section that requires testing of the heat-affected zone. Charpy V-notch impact tests are performed on the base metal and on the HAZ at the same temperature. The results are acceptable if the lateral expansion of the HAZ specimens equals or exceeds the lateral expansion of the base metal. If the average lateral expansion of the HAZ specimens is less than that of the unaffected base metal, either of two avenues may be followed. The welding procedure qualification may be repeated. Alternatively, the temperature at which the average lateral expansion of the HAZ equals the lateral expansion of the unaffected base material is determined; the number of degrees by which this temperature exceeds the base-metal testing temperature is added to either:

1. RT_{NDT} for material covered by NB-2331, "Material for Vessels."

2. The lowest service temperature (LST) for material covered by NB-2332, "Material for Piping Pumps, and Valves, Excluding Bolting Material."

This additive temperature would be applied to all material welded with this weld procedure. The remainder of NB-4340 contains requirements for certain specific types of welds, (i.e., tube-to-tubesheet and specially designed seals).

Pressure Vessels Fabricated by Welding

Service requirements for vessels under Section VIII of the code, Part UW, are classed by the type of service intended: containment of lethal substances, operation below -20°F (-29°C), and fired or unfired steam boilers. The type of service establishes special requirements for the types and degree of inspection of welded joints. Materials used in Part UW vessels must meet the standards of Section IX, "Welding and Brazing Qualifications." Additional requirements by class of material are incorporated in Subsection C of Section VIII.

Part UHT of Section VIII is "Requirements for Pressure Vessels Constructed of Ferritic Steels With Tensile Properties Enhanced by Heat Treatment." All steels used in Part UHT vessels are tested for notch ductility. These tests are performed at the lowest temperature at which pressure is applied to the vessel, or at the design temperature, whichever is lower, but in no case can the test temperature be higher than 32°F (0°C).

All test specimens are taken from full-thickness samples in their final heat-treated condition. When the material is clad or weld-overlaid prior to quenching and tempering, the full-thickness samples are to be similarly clad or weld-overlaid before heat treatment.

One Charpy V-notch test is to be made on each of three specimens from each plate as heat treated and from each heat of bar, pipe, forgings, or castings included in one heat-treatment lot. Each specimen must exhibit at least 15 mils lateral expansion. If the specimens do not exhibit the necessary ductility, the part may be re-heat treated and retested. An additional requirement is specified for SA-517 and SA-592 (Cr-Mo and Cr-Mn-Ni steels) for use at design temperatures below -20°F (-29°C) and for SA-645 (Ni-Mo steel) for use below -285°F (-171°C). For these materials, one drop-weight test (two specimens; ASTM E-208) should be conducted for each plate 5/8 in. or over as heat treated and for each heat of each heat-treatment lot for forgings and castings of all thicknesses.

Testing after heat treatment, not merely in simulated conditions, is required. Coupons representative of the material and welding in each vessel are heat treated with the vessel or component. One tensile and one impact test are required for each lot of material represented in the vessel or component. All welding procedures and welders must be qualified to the requirements of Section IX.

Section VIII, Division 2, contains the only exception to the matching strength philosophy. The 9 percent nickel grades included in SA-353, SA-333, and SA-334 are allowed to have a weld-qualification tensile strength of 95 ksi, 5 percent below the minimum, for cryogenic applications. However, where these steels are to be used at elevated temperatures as specified in Section VIII, Division 1, a penalty must be taken for the reduced weldment strength.

Summary

The use of a specific filler metal is not dictated by any portion of the B&PV Code, and yield-strength requirements are not mentioned. What is required is a demonstration that the weldments meet the minimum requirements for tensile strength (with the exception of SA-522 9 percent Ni steel), ductility, and toughness of the base metal. The Code, therefore, would place no restriction on using an austenitic stainless steel filler metal (yield strength = 30 ksi, tensile strength = 80 ksi) to weld a medium-strength low-alloy steel (yield-strength = 50 ksi, tensile strength = 80 ksi).

The Code's design philosophy provides an explanation for this approach. Relatively, considerably more analysis and, therefore, less conservatism, in the form of safety factors, distinguish Section III (Nuclear Power Plant Components), which thus will be used as the primary example of the code's philosophy.

Section III uses as a design criterion the maximum shear stress or Tresca theory because it is relatively simple and somewhat more conservative than the distortion energy or Von Mises theory. In applying this criterion to the design of structures, four basic categories of stresses have been designated and allowable limits determined for them in terms of yield strengths. However, to prevent unsafe designs with respect to high-strength materials with low ductility and high yield strength ratios, the ultimate strength of a material is considered in determining the stress limits. For this reason, the stress analysis performed for primary components is based on "stress intensity," S_m, that is, the allowable primary membrane stress. It is defined as less than or equal to two thirds the yield strength (2/3 S_y) or one third the ultimate strength (1/3 S_u), whichever is less. Therefore, with materials comparable to HY-80, HY-100, and HY-130, where the yield-to-ultimate strength ratio is greater than 0.5, i.e., the S_y to S_u ratio, it is the ultimate strength and not the yield strength that defines the stress limit (ASME, 1964).

In general, the B&PV Code considers weld metal to be homogeneous with respect to properties when compared to base metal. It can be seen, therefore, why ultimate tensile strength is the criterion for weld-procedure qualification. Very few allowable materials exhibit an S_y/S_u ratio of less than 0.5; therefore, the weldment will exhibit the same allowable stress, S_m, as the base metal, regardless of the weld metal's yield strength.

As noted earlier, the ASME Boiler and Pressure Vessel Code makes only one exception to its overall philosophy of matching weld-metal strength and toughness with those of the base metals being joined. That is the 5 percent reduction in allowable stress for 9 percent nickel steel.

Structural Welding Code, AWS D1.1

The AWS Structural Welding Code D1.1 provides requirements for the welding of structures such as bridges and buildings. It is referenced generally by the design specification for the particular structure. Similar to the ASME B&PV Code, D1.1 contains general requirements for preparing and qualifying welding procedures and has supplemental requirements for specific structures; such as existing structures, new buildings, new bridges, and new tubular structures.

The weld-metal strength required by the American Welding Society (AWS) D1.1 are based on the type of loading on the weld joint (Table 5-1). For complete-penetration groove welds stressed in tension normal to the effective area, matching weld metal is required. For geometries where the stress is not normal to the effective area, or is compressive, weld metal of reduced strength may be used.

Materials

The materials allowed for welding construction in AWS D1.1 are basically carbon and low-alloy steels that have minimum yield strengths of up to and including 100 ksi. Examples of these high-strength materials are ASTM A-514, ASTM A-517 (Cr, Mo, and Cr, Mo, Ni steels) and ASTM A709 Grades 100, 100W (HSLA steel). Filler metal must conform to various AWS filler metal specifications requiring yield and tensile strengths that range up to that of the base metals.

Weld Qualification

Qualified welding procedures used in accordance with D1.1 are similar to those described for the ASME B&PV Code. A major difference is that requalification is not required when changing to a lower-strength filler metal because design philosophy differences between the two permit this variance. Specifically, the B&PV Code regards the weld metal and base metal as homogenous with respect to mechanical properties.

American Petroleum Institute (API)

A number of API standards for welded construction were reviewed for their requirements for filler metal strength. These cover the fabrication of welded tanks for refineries (API 620 and API 650) and welded pipelines for both liquid and gas (Code of Federal Regulations, Title 49, Parts 191, 192, and 195).

API does not place specific controls on the use of filler metals. However, the governing body does share the philosophy that the welding material must be equal to or better than the base material. The single exception to this requirement, as in the ASME B&PV Code, is the 9 percent nickel steels for cryogenic storage. For this case, slight undermatching of ultimate strength at room temperature is allowed. The exception is based on the extreme toughness of these materials and the increase in their yield strength with decreasing temperature.

API pipeline construction is controlled by the Code of Federal Regulations, Title 49, Part 195.220. The code states, "Filler metal must be equal in strength to the highest specified minimum yield strength of the member being welded..." (Private communication with R. Griffin, Chairman of the API Committee, July, 1980).

TABLE 5-1 Allowable Stresses in Welds

<table>
<tr><th>Type of weld</th><th colspan="2">Stress in weld</th><th>Allowable stress</th><th>Required weld strength level</th></tr>
<tr><td></td><td colspan="2">Tension normal to effective area</td><td>Same as base metal</td><td>Matching weld metal must be used.</td></tr>
<tr><td rowspan="3">Complete joint penetration groove welds</td><td colspan="2">Compression normal to effective area</td><td>Same as base metal</td><td>Weld metal with a strength level equal to or one classification (10 ksi) less than matching weld metal may be used.</td></tr>
<tr><td colspan="2">Tension or compression parallel to axis of the weld</td><td>Same as base metal</td><td rowspan="2">Weld metal with a strength level equal to or less than matching weld metal may be used.</td></tr>
<tr><td colspan="2">Shear on effective area</td><td>0.27 nominal tensile strength of weld metal (ksi), except shear stress on base metal shall not exceed 0.36 yield strength of base metal</td></tr>
<tr><td rowspan="5">Partial joint penetration groove welds</td><td rowspan="2">Compression normal to effective area</td><td>Joint not designed to bear</td><td>0.45 nominal tensile strength of weld metal (ksi), except stress on base metal shall not exceed 0.55 yield strength of base metal</td><td rowspan="5">Weld metal with a strength level equal to or less than matching weld metal may be used.</td></tr>
<tr><td>Joint designed to bear</td><td>Same as base metal</td></tr>
<tr><td colspan="2">Tension or compression parallel to axis of the weld</td><td>Same as base metal</td></tr>
<tr><td colspan="2">Shear parallel to axis of weld</td><td>0.27 nominal tensile strength of weld metal (ksi), except shear stress on base metal shall not exceed 0.36 yield strength of base metal</td></tr>
<tr><td colspan="2">Tension normal to effective area</td><td>0.27 nominal tensile strength of weld metal (ksi), except shear stress on base metal shall not exceed 0.60 yield strength of base metal</td></tr>
<tr><td rowspan="2">Fillet welds</td><td colspan="2">Shear on effective area</td><td>0.27 nominal tensile strength of weld metal (ksi), except shear stress on base metal shall not exceed 0.36 yield strength of base metal</td><td rowspan="2">Weld metal with a strength level equal to or less than matching weld metal may be used.</td></tr>
<tr><td colspan="2">Tension or compression parallel to axis of weld</td><td>Same as base metal</td></tr>
<tr><td>Plug and slot welds</td><td colspan="2">Shear parallel to faying surfaces (on effective area)</td><td>0.27 nominal tensile strength of weld metal (ksi), except shear stress on base metal shall not exceed 0.36 yield strength of base metal</td><td>Weld metal with a strength level equal to or less than matching weld metal may be used.</td></tr>
</table>

SOURCE: American Welding Society D.1.1 Structural Welding Code.

Conclusions

The welding standards examined here represent those used in the United States to design and fabricate a wide variety of systems and structures. This review of the codes, conducted to find precedents related to a more effective use of the yield strength of welding filler materials revealed two basic points: for conservatism, the design of weldments in these structures is based generally on some percentage of the ultimate tensile strength of the joint or filler metal; and for the majority of applications, the codes require a weld joint at least equal in strength to that of the base metal. However, code precedence exists for undermatching weld-metal strength where the stresses are not oriented in critical directions. A special case of limited undermatching also exists for a specific type of high-toughness material where toughness is of primary importance.

REFERENCES

American Society of Mechanical Engineers, Boiler and Pressure Vessel Code, Section III, Nuclear Power Plant Components, New York, N.Y.; American Society of Mechanical Engineers, 1980.

American Society of Mechanical Engineers, Boiler and Pressure Vessel Code, Section VIII, Pressure Vessels, New York, N.Y.; American Society of Mechanical Engineers, 1980.

American Society of Mechanical Engineers, Boiler and Pressure Vessel Code, Section IX, Welding and Brazing Qualification, New York, N.Y.; American Society of Mechanical Engineers, 1964.

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U.S. Department of Transportation, Code of Federal Regulations, Title 49, Parts 191 and 192 Regulations for the Transportation of Natural Gas and Other Gases by Pipeline, Washington, D.C.; 1979.

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CHAPTER 6

HYDROGEN-INDUCED CRACKING

The ability of hydrogen to cause cracking in the weldments of hardenable steels is well known. Hydrogen cracking may occur in either the heat-affected zone (HAZ) or the weld metal if four conditions are present simultaneously. These conditions have been defined as (Sawhill et al. 1974):

1. A critical concentration of diffusible hydrogen at a crack tip.
2. A stress intensity of significant magnitude.
3. A temperature in the range of -150°F to 400°F (-101°C to 204°C).
4. A susceptible microstructure.

Although the conditions necessary for hydrogen-induced cracking are well-known, there exists considerable controversy over the mechanism of hydrogen cracking. Four basic hypotheses have been proposed to explain this phenomenon.

Hydrogen Cracking Hypotheses

The essential points of the four basic hypotheses that describe possible mechanism for hydrogen-induced cracking are presented below. A critique is offered at the end of this section to identify the strengths and weaknesses of each.

Planar-Pressure Mechanism

The planar-pressure mechanism of hydrogen cracking was proposed initially by Zapffe and Sims (1941) and later modified by Tetelman (1962). It primarily relates to the variation in solubility with temperature of hydrogen in iron (Figure 6-1). Note that hydrogen solubility falls abruptly from about 25 ppm in liquid iron to about 6 ppm in body-centered cubic delta ferrite at the melting point (1538°C). The solubility increases from about 4.9 to 8.3 ppm when delta ferrite transforms to face-centered cubic austenite at 1394°C. The solubility in austenite decreases continuously to about 4.1 ppm at 912°C and then drops to 2.7 ppm when the austenite transforms to ferrite at 912°C. The solubility in ferrite then decreases to less than 0.5 ppm at ambient temperature.

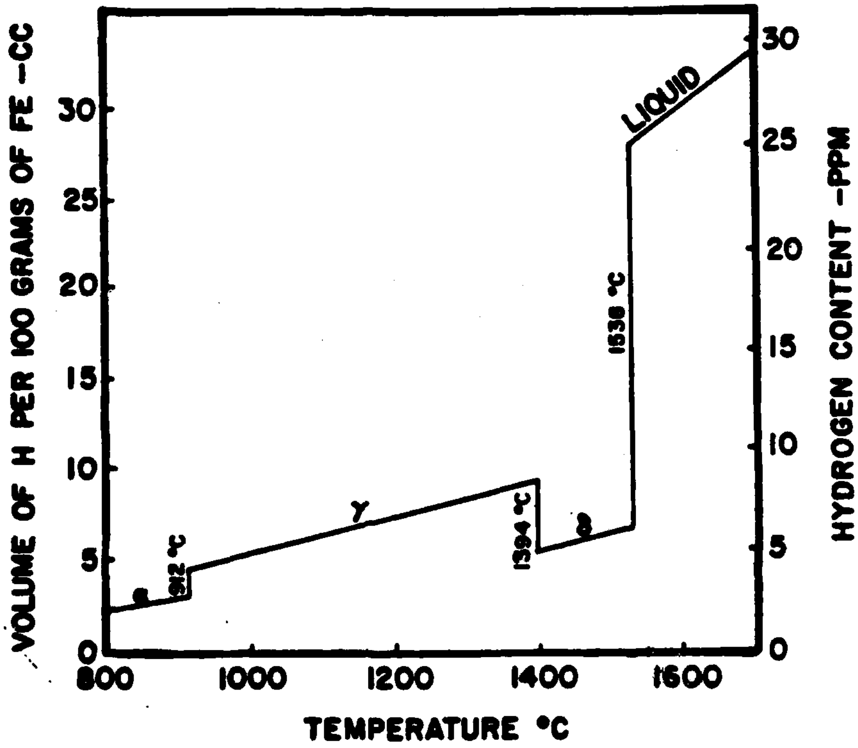


Figure 6-1 The solubility of hydrogen in iron at one atmosphere pressure. (Note: Hydrogen concentration, in cubic centimeters per 100 g of iron is equal to 1.12 times the concentration in parts per million.)
Note: Hydrogen concentration, in cubic centimeters per 100 g of iron is equal to 1.12 times the concentration in parts per million.

Source: Tetelman 1962.

This hypothesis postulates that atomic hydrogen diffuses through the iron lattice and precipitates in voids or other defects. Inside these voids, the reassociated atomic hydrogen, as H_2 gas, produces an extremely high internal pressure. The interaction of high internal pressure with the existing applied and residual stresses is postulated to result in crack initiation and propagation from the initial defect sites.

Surface-Adsorption Hypothesis

The surface-adsorption hypothesis was proposed initially by Petch (1956) and recently modified by Williams and Nelson (1970). This hypothesis postulates that hydrogen adsorbed on the surfaces of internal cracks lowers the surface free energy and thus reduces the critical fracture stress. Thus, the initial crack is formed by dislocation pileups at a grain boundary, and hydrogen is postulated to influence crack propagation, but to have no effect on crack initiation.

Triaxial-Stress Hypothesis

Troiano (1960) has advanced a triaxial-stress hypothesis which assumes that hydrogen in solution near a void causes embrittlement. This hypothesis predicts that the concentration of hydrogen in the region of maximum triaxial stress near a void or stress raiser exceeds that in the lattice at equilibrium. It is assumed that hydrogen in the void itself is nondamaging and that a critical stress must be exceeded to cause a critical or damaging concentration of hydrogen.

According to Troiano, the diffusion of hydrogen is induced by the stress gradient created by the region of triaxial stress. When the hydrogen reaches a critical concentration in the region of maximum triaxial stress, a crack is initiated. The crack is propagated by this same mechanism after hydrogen diffuses to the triaxial-stress region ahead of the newly created crack tip.

Hydrogen-Dislocation-Interaction Hypothesis

The fourth hypothesis of hydrogen cracking stresses the interaction of hydrogen with moving dislocations. Two opposing hypotheses of this interaction have been advanced. Beachem (1972) proposed a hypothesis involving microplasticity mechanisms instead of embrittlement mechanisms. In this model, the pressure of hydrogen dissolved in the lattice ahead of the crack tip assists microscopic deformation processes. The presence of dissolved hydrogen, then, lowers the applied stress necessary to cause fracture by a dislocation-multiplication mechanism.

On the other hand, Steinman and coworkers (1965) and Bernstein (1965) suggest that hydrogen in the lattice reduces the energy required to initiate and propagate fracture by reducing dislocation mobility. This is postulated to cause the development of large local strain concentrations that lead to the initiation and propagation of cracks.

Critique of Hypotheses

Planar-Pressure Hypothesis: Blisters containing molecular hydrogen have been noted in steels that have been cathodically charged with hydrogen or have been exposed to certain corrosive media. In these cases, atomic hydrogen, absorbed by the steel, has diffused interstitially to voids where it recombined to form molecular hydrogen. Under high internal pressure, these voids deform to become blisters in ductile steels.

However, it has been shown by Johnson (1969) that steels of high yield strength are embrittled more by one atmosphere of gaseous hydrogen than by electrolytic charging. Because the pressure-expansion mechanism cannot operate in the case of diffusion under a single atmosphere, the mechanism of hydrogen embrittlement cannot be satisfactorily described by the planar-pressure hypothesis. Undoubtedly, this mechanism will contribute stresses when the thermodynamic activity of hydrogen is very high. However, in the case of hydrogen-induced cracking in steels of high yield strength, only a few parts per million of hydrogen are required.

Surface-Absorption Hypothesis: The surface-absorption hypothesis of Petch (1956) and the modified versions of Williams and Nelson (1970) have been criticized on the basis of the fact that the incubation time for crack initiation is reversible with respect to the applied stress. Furthermore, the embrittlement persists at low temperatures, where both the bulk diffusion rate and the rate of redistribution of hydrogen along newly created surfaces would be negligibly small (Troiano 1960).

Hydrogen-Dislocation-Interaction Hypothesis: The model of hydrogen-dislocation-interaction by Beachem (1972) seems to unify satisfactorily several hypotheses of hydrogen-induced cracking. The model suggests that the presence of a sufficient concentration of hydrogen dissolved in the lattice just ahead of the crack tip aids whatever deformation processes the microstructure will permit.

Experimental investigations indicate that hydrogen-induced cracking (HIC) is both delayed and discontinuous (Savage et al. 1976a, b). The augmented strain-cracking test (Savage et al. 1976b) was utilized to observe microscopically crack propagation and the associated evolution of hydrogen. Time-lapse motion pictures (Nippes 1975), taken at two frames per second, document the principal features of crack propagation such as the nucleation of microcracks, the development of localized yielding ahead of propagating cracks, and the evolution of hydrogen bubbles from cracks. The film demonstrates that the v-shaped zone of localized yielding ahead of the propagating crack becomes larger as the amount of diffusible hydrogen is increased. Crack growth is intermittent and hydrogen evolution occurred during crack-arrest periods. The intermittent hydrogen evolution occurred not at the crack tip, but from the crack walls some distance behind the crack tip.

These experiments confirm that hydrogen diffuses toward the triaxially stressed crack tip until a critical concentration is reached for the stress involved. The crack then moves rapidly until the hydrogen concentration ahead of the crack becomes too low for further propagation of the crack. Local stress release causes hydrogen gas to evolve from the newly formed walls of the crack. These intermittent processes continue until the hydrogen concentration or the stress becomes too low for hydrogen-induced cracking to continue.

Control of Hydrogen-Induced Cracking

A discussion of various factors that control HIC are presented below.

Hydrogen Level

During welding, hydrogen gas from the arc atmosphere is converted to the atomic state and dissolved readily in the weld pool. Because its solubility in steel decreases with decreasing temperature, there is a strong driving force for the rejection of hydrogen from the HAZ and weld metal during cooling. To escape, atomic hydrogen must diffuse to some interface, collect, and reform as molecular hydrogen. However, it is postulated that atomic hydrogen may interact with dislocations and diffuse to triaxially stressed regions where it acts as an embrittling agent. Further, because it is very difficult for atomic hydrogen to escape from lattice imperfections by diffusion, extremely high internal stresses can develop and cracking may occur.

The amount of hydrogen absorbed by a weldment depends on several factors including: the time and temperature conditions of cooling, the size of the weld bead, and the initial concentration of hydrogen in the arc atmosphere. Generally, the risk of cracking increases with increasing hydrogen concentration. Control of the hydrogen level may be achieved by minimizing the available hydrogen and providing sufficient time for hydrogen to diffuse from the weldment. The major sources of hydrogen are hydrogenous compounds and moisture in the coatings of SMAW electrodes, contamination in the shielding gas, contamination of bare filler wire, and surface contamination of the base metal.

The potential hydrogen level of a process is determined by measuring the moisture or hydrogen content of the welding consumables. The higher the potential hydrogen level, the higher will be the hydrogen content of the weld. Although potentially available, not all the measured hydrogen is absorbed by the weld pool (Coe 1973) (Figure 6-2).

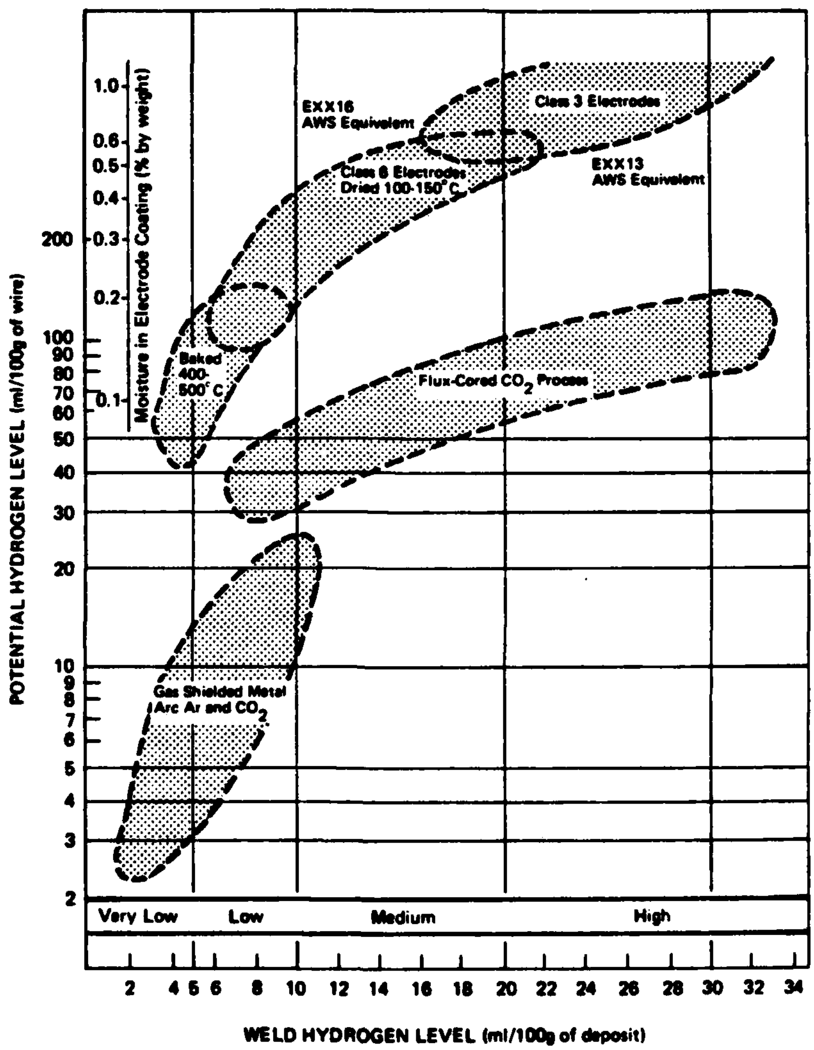


Figure 6-2 Available hydrogen related to the weld metal by hydrogen level.

Source: Coe 1973.

Control of the hydrogen potential of most consumables depends on the conditions under which they are stored (DeLong, 1968). In general, the potential hydrogen level of electrode coatings and fluxes can be decreased by lowering the moisture content by drying or baking. However, efforts to eliminate hydrogen cracking in HY-80 by controlling the hydrogen content in bare filler wire have not been totally successful. The specifications for welding consumables for HY-80 call for a maximum of 0.2 percent moisture in electrode coatings and 6 to 8 ppm hydrogen in bare filler wires. Unfortunately, HY-130 weldments are even more susceptible to hydrogen-induced cracking than are HY-80 weldments. The more stringent specifications for HY-130 welding consumables require a maximum of 0.1 percent moisture in electrode coatings and 3 ppm hydrogen in bare wires (Gross 1968) (Figure 6-3).

Such requirements for controlling the hydrogen content of arc atmospheres are not easily met in a shipyard environment. It therefore has become necessary to study other procedures for eliminating hydrogen-induced cracking. One such procedure attempts to render weld- metal hydrogen innocuous by chemical combination to form a hydride (Howden 1976). Rare earth metals can be added to the filler metal to form hydrides.

Unfortunately, such additions have been shown to make the manufacturing of filler wire more difficult and possibly might be detrimental to the mechanical properties of HY-130.

The implant test (Figure 6-4) is used for studying hydrogen-induced cracking (Granjon 1969, 1973). The limiting cracking stress (LCS) for a particular head of steel depends on the hydrogen content of the weld. A HY-80 heat, welded with three different SMAW electrodes yielded different levels of diffusible hydrogen (Howden 1981):

	Hydrogen content, ml/100g,		LCS
Electrode	Fused Metal	Deposited Metal	ksi
E-11018-M, Moist. Resist.	1.76	1.00	88.0
E-11018-M, Regular Supply*	3.48	1.92	59.0
E-9010-G	50.20	26.00	33.0

* Electrodes baked for one hour at 800°F immediately prior to welding.

These implant test results, plotted in Figures 6-5 to 6-7, show that the value of LCS decreases rapidly as the hydrogen content increases.

Experiments with the same heat of HY-80 steel (Howden 1981) demonstrated the effect of preheat on the LCS value. Figure 6-8 shows that increasing the preheat temperature, which promotes hydrogen diffusion from the weld, is very effective in raising the LCS value and lowering the susceptibility of a steel to hydrogen-induced cracking.

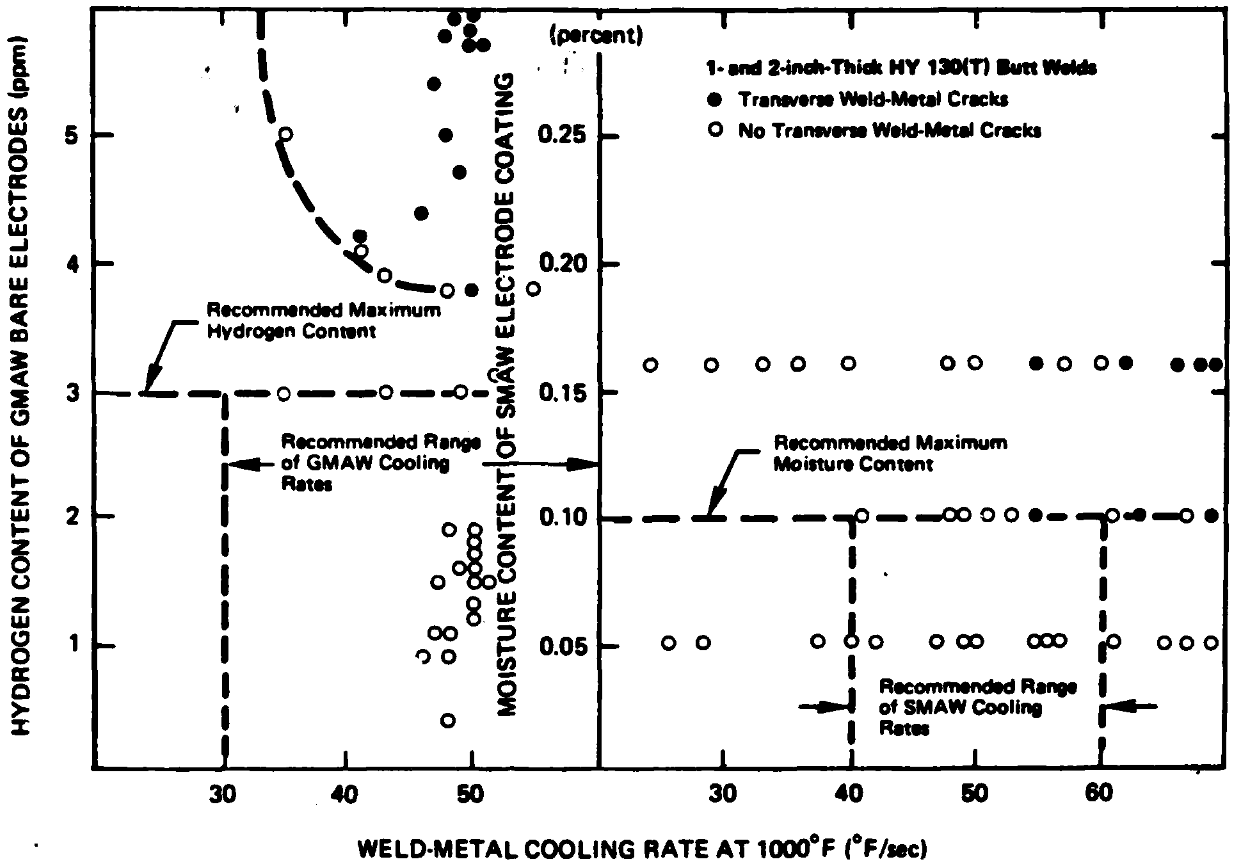


Figure 6-3 Effect of hydrogen in gas metal-arc (GMA) bare electrodes and moisture in shielded metal-arc (SMA) electrode coatings on weld metal cracking.

Source: Dorschu and Lesnewich 1964.

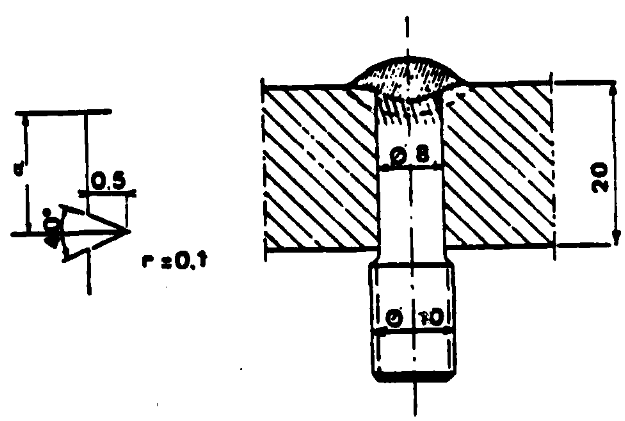


Figure 6-4 Specimen for cold-cracking test.

Source: Masubuchi 1980.

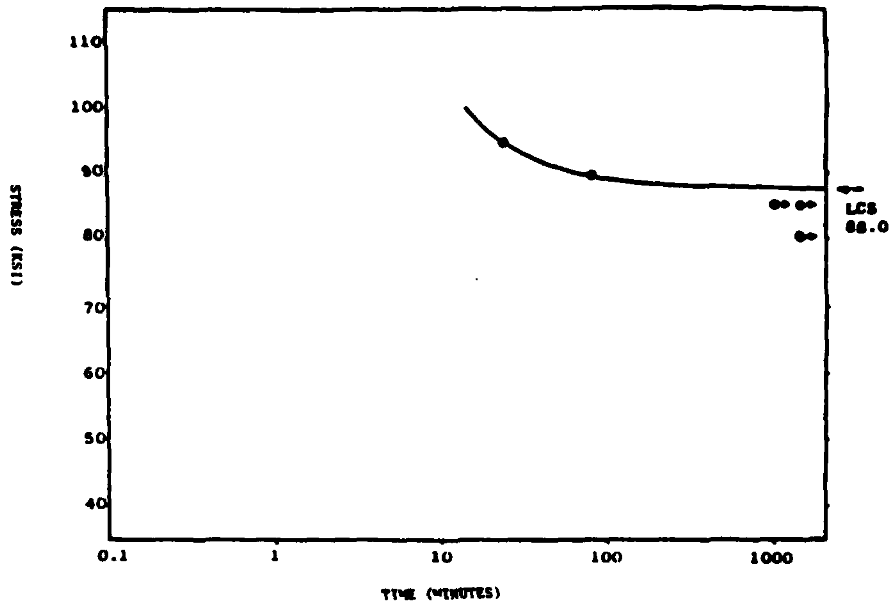


Figure 6-5 Stress-time plot for HY-80 steel E11018-M moisture resistant electrode, 78°F preheat.

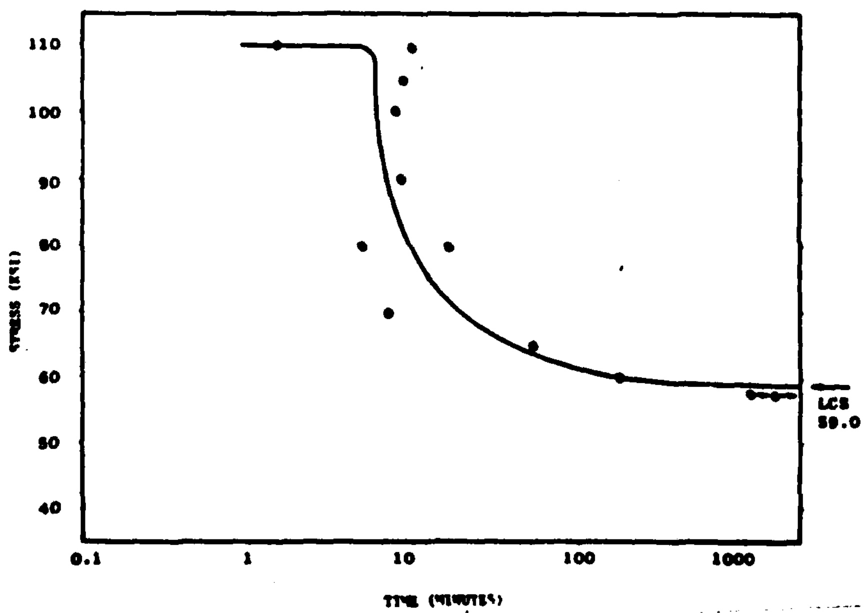


Figure 6-6 Stress-time plot for HY-80 steel, E11018-M electrode, 78°F preheat.

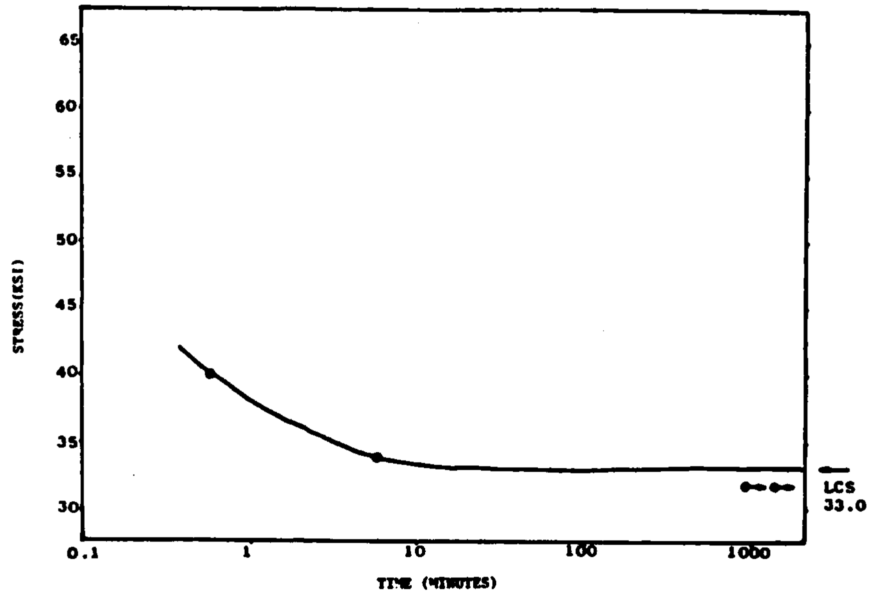


Figure 6-7 Stress-time plot for HY-80 steel, E9010-G electrode, 78°F preheat.

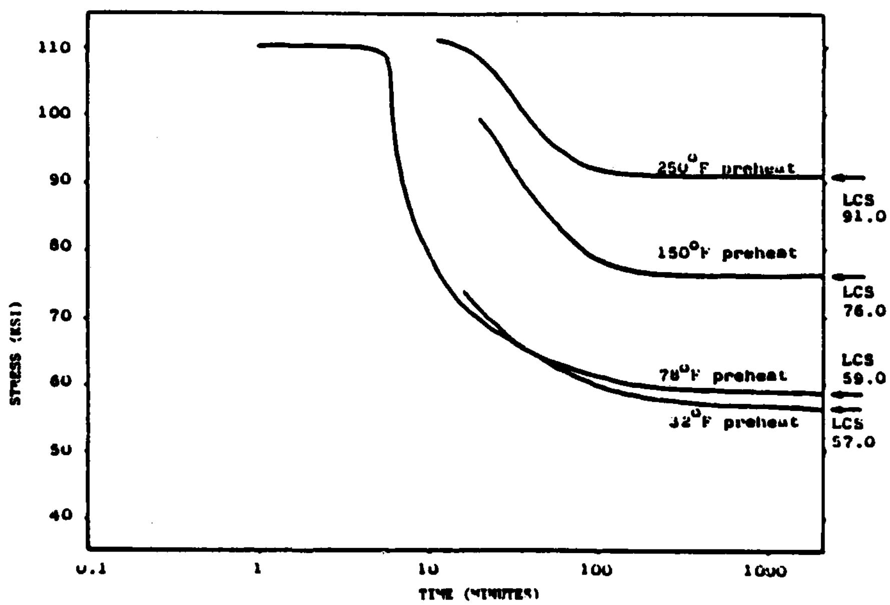


Figure 6-8 Stress-time plot for HY-80 stee, E11018M electrode, 32°F, 78°F, 150°F, and 250°F preheat.

Stress Level

Welding stresses arise from three factors: external restraint of the welded sections, unequal thermal expansion and contraction of the base metal and weld metal, and volumetric expansion resulting from microstructural changes in a weldment. The stresses on a weld are a function of weld size, joint geometry, fixturing, welding sequence and the yield strengths of the base and weld metals. Of these factors, it has been found that poor joint fit-up (root gaps greater than 0.4 mm) markedly increases the chance of hydrogen-induced cracking (Coe 1973).

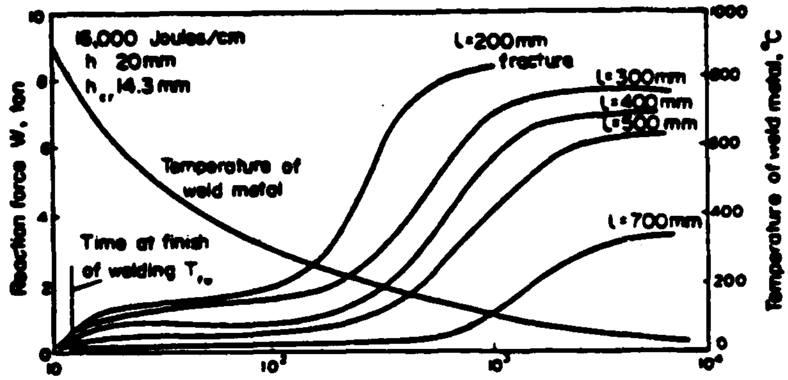
Hydrogen introduced into a weldment lowers the stress level at which cracking will occur by reducing the cohesive strength of the lattice and by adding to the localized stresses at discontinuities. The stresses generated interact with hydrogen to enlarge discontinuities into cracks. To prevent cracking, the stresses developed must be accommodated by strain in the weld metal. Selecting the lowest strength weld metal allowable by the design, in conjuction with good welding practice, will reduce weld stress and therefore the probability of hydrogen-induced cracking (Coe 1973).

The rigid restraint cracking (RRC) test has been used (Watanabe 1964; Satoh 1967, 1975) as well as the implant test to study the effect of restraint stress on hydrogen-induced cracking. The reaction force increases rapidly when the restraint length is small and could become high enough shortly after completing the weld to crack the weld metal. A larger restraint length causes the reaction force to increase more slowly and thus cracking is delayed. Large values of restraint length will never raise the reaction force high enough to cause cracking. The effect of restraining gage length on reaction force for a low-carbon steel is shown as a function of time in Figure 6-9. Figure 6-10 also shows the temperature-time relationship of the weld metal. These experiments showed that cracking occured with a restraint length of 200 mm, but not at 300 mm. Typical results for HT-80* in Figure 6-10 show the development of restraint stress during cooling and cracking in the RRC test. Small values of restraint length cause restraint stresses build up quickly after welding to produce hydrogen-induced cracking. As the restraint length is increased, the restraint stresses build up slowly thus delay cracking. Hydrogen-induced cracking does not occur below a critical value of restraint stress. Satoh (1975) found that the critical stresses for crack initiation in the RRC correlate very well with the implant test results. Figure 6-11 shows such comparative data involving seven heats of steel, including two cases having a preheat of 50°C (122°F).

Critical Temperature Range

After the application of a stress, a specific time delay for the appearance of hydrogen-induced cracks is required. This time delay, which is somewhat dependent on the stress, decreases with increasing hydrogen

*HT is the Japanese designation for high tensile strength steels; it is similar to but not directly equal to the U.S. HY designation.



Time after start of welding, sec

Figure 6-9 Effect of restraining gage length (l) on reaction force.

Note: Low-carbon steel weldments, plate thickness h=20mm, σ = 16,000 joules/cm, breadth of specimen (weld lenth) = 30 mm.

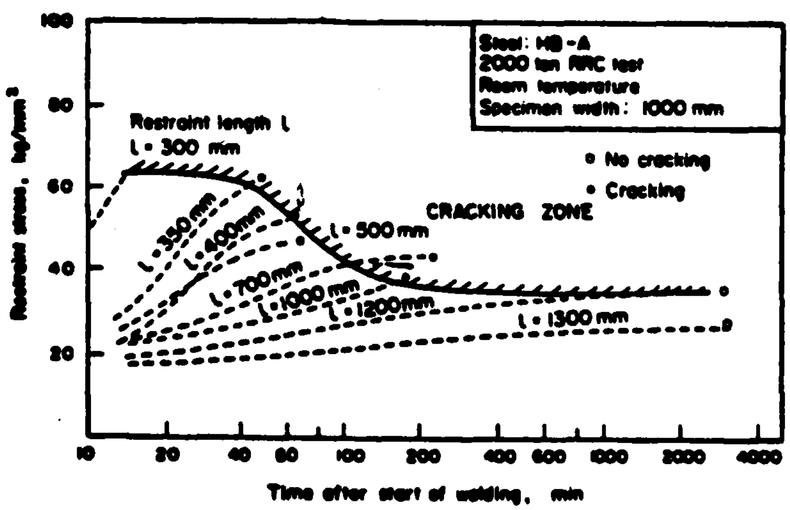


Figure 6-10 Development of restraint during cooling and cracking in the RRC test.

Source: Masubuchi 1980.

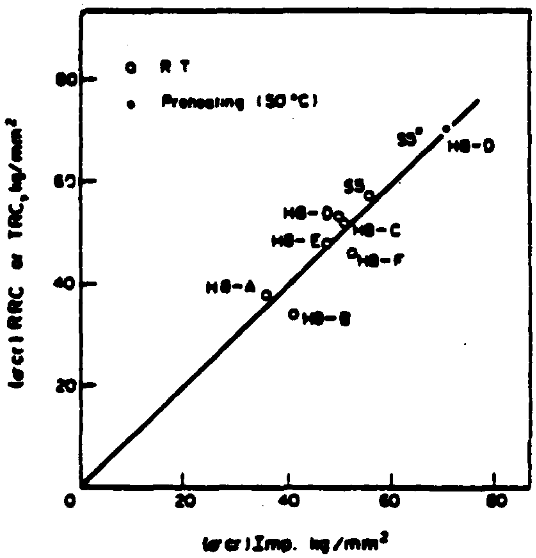


Figure 6-11 Relation between σ_{cr} of implant test and that of RRC and TRC tests.

Source: Masubuchi 1980.

concentration and with increasing strength of the steel. The time delay in hydrogen-induced cracking results from the time required for the hydrogen diffusion to produce the critical hydrogen concentration at the crack tip. Below -150°F (-101°C) the diffusion of hydrogen is so slow that hydrogen-induced cracking is minimized. On the other hand, the rate of diffusion of hydrogen above 400°F (204°C) is so fast that it is impossible to accumulate the critical hydrogen concentration at the crack tip.

Type of Microstructure

Microstructure is an important factor in the susceptibility of a steel to hydrogen-induced cracking. The microstructure can provide a crack-sensitive matrix that can operate with destructive results in the presence of hydrogen. Any microstructure that possesses low ductility and contains internal stresses will be sensitive to hydrogen-induced cracking. Generally, the higher the carbon content and the harder the microstructure, the more susceptible the steel will be to hydrogen-induced cracking. The following microstructures are in increasing order of susceptibility to hydrogen-induced cracking (Graville 1968).

- (a) Nonmartensitic structures such as bainite or ferrite.
- (b) Mixtures of martensite and some softer phase, such as ferrite or bainite.
- (c) Martensite.
- (d) Martensite with internal twinning.

The use of a large weld bead, thin plate, high heat input, and high preheat will reduce the quenching rate in the HAZ and provide a softer, less sensitive microstructure. The HAZ may also be softened either as a result of a stress relief heat treatment or the tempering effect of subsequent weld passes.

Thermal Treatment

It is possible to reduce the susceptibility of a weldment to hydrogen-induced cracking by adjusting process parameters and sequences. DeLong (1968) recommends three simple rules to follow in joining steels with yield strengths of approximately 140 ksi. These rules were found to apply to the welding of all steels. The basic rules are: 1) deposit thin beads, 2) provide for a delay time between the deposition of successive passes and closely control the interpass temperature, 3) maintain the interpass temperature for a reasonable length of time after completing the weld (one hour is generally the minimum time allowance). These procedures promote the diffusion of hydrogen from the weld.

In weld thermal cycles, the magnitude of the HAZ cooling rate which occurs at 1000°F (540°C) is important in determining the austenite transformation products. However, the magnitude of the cooling rate at 1000°F (540°C) is not as important in determining the amount of diffusible hydrogen liberated from the weld and the HAZ as is the cooling rate at 400°F (200°C) and lower. Two welds having *identical cooling rates at* 1000°F (540°C)--one made with higher energy input, no preheat, and low interpass temperature; the other with a lower energy input and a 300°F (150°C) preheat and interpass temperature--with the preheat and interpass temperature the latter weld would have a lower cooling rate at 400°F (200°C). This lower cooling rate at slower temperatures would provide a longer time for hydrogen to diffuse from the weldment.

DeLong (1968) has also recommended a minimum postheat time of one hour and a temperature of 275°F (135°C) for welds made with E14018 electrodes. Graville (1968) has suggested a minimum time of one half hour for postheats above 480°F (250°C) and increasing times for postheats at lower temperatures. As might be expected, both postheat time and temperature depend on weld bead thickness (Interrante and Stout 1968). Today's specifications for the fabrication of HY-80 recommend a preheat and postheat treatment of 200°F (90°C) (minimum) to 300°F (150°C) (maximum) (Robbins 1960; Flax et al. 1971). Such heat treatments are both cumbersome and expensive.

The D. W. Taylor Naval Ship Research and Development Center (DTNSRDC) has been investigating postweld heat treatments of the HY-80 and HY-130 systems. Figure 6-12 summarizes the restrained butt weld data subsequently developed by the shipyards showing the various temperature-time thermal soaking combinations as a function of the ductility (percent elongation) at failure under a tensile load (MIL-STD-1681, 1976). The data presented are for thermal soaking at 1/2 in. (1.27 cm) thick increments (intraweld), thermal soaking after each weld layer is completed (interlayer), and after the total weld is

finished (postweld thermal soaking). The HY-80 system does not require a postweld treatment to meet specification requirements. It was observed that HY-80 restrained welds exhibited a loss of ductility when produced without intermediate thermal treatment. However, when the HY-80 restrained welds were subjected to a 250°F (120°C) postsoak for 12 h, the ductility was satisfactory.

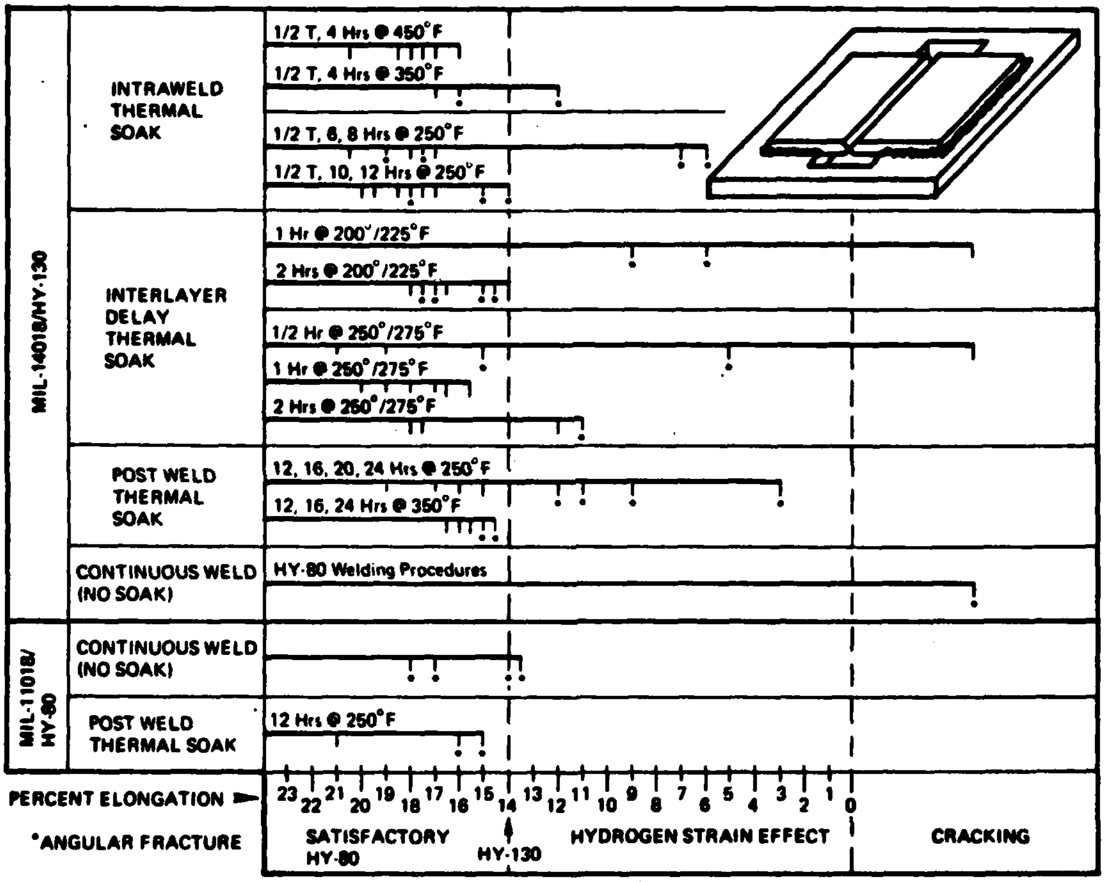


Figure 6-12 Effect of thermal treatment on hydrogen strain damage of restrained butt welds.

Source: DTNSRDC 1974.

In general, longer soak times and higher soak temperatures were more successful in eliminating hydrogen-strain damage. For example, intraweld soaking of four hours at 450°F (230°C) produced acceptable welds, whereas lowering the soak temperature to 350°F (180°C) for four hours produced one unacceptable tensile test out of three. Lowering the temperature to 250°F (120°C) and increasing the soak time to 10 and 12 h produced acceptable welds as measured by the 14 percent elongation requirement, although three out of eights welds tested had angular-type fractures.

Because interlayer and intraweld thermal soaking are impractical under present field conditions, postweld soaking at 350°F (180°C) for 12 h (Figure 6-12) is considered the best way of relieving the hydrogen-strain effect under shipyard conditions. MIL-14018 requirements for removing hydrogen from weld metal, currently defined in the Fabrication Document (MIL-STD-1681, 1976), are shown in Table 6-1.

In another DTNSRDC program, several thermal soaking treatments were investigated to find one applicable to removing damaging hydrogen from repair weldments (Juers 1975). HY-130 trough weldments (Figure 6-13) were fabricated for this purpose using the SMAW process; 1/8-in. MIL-14018 electrodes were evaluated under a number of thermal conditions (Figure 6-14). Continuous welding with no thermal soak demonstrated the weld metal's propensity to crack (see bottom portion of Figure 6-14). Thermal treatments included interlayer, intraweld, and postweld soaking at normal fabrication temperatures and at elevated temperatures. Only the soaking treatments at elevated temperature (400°F) resulted in consistently acceptable recovery of ductility in the weld metal
(Juers 1975).

Yield Strength and Hydrogen Cracking

When hydrogen-induced cracking occurs in weldments of mild C-Mn and low-alloy steels with yield-strengths lower than 85,000 psi, generally cracking is located in the heat-affected zone rather than in the fusion zone. However, under certain conditions, such as those involving high stress concentrations in the fusion zone hydrogen-induced cracking may be encountered in the weld metal. This latter cracking almost always originates in the heat-affected zone and propagates into the weld metal. In these cases, when the welding parameters are adjusted so that hydrogen-induced cracking in the heat-affected zone is eliminated, the cracking in the weld metal is also eliminated.

The situation is different, however, when steels with yield strengths greater than 85,000 psi are joined with weld metal of matching or overmatching yield strength. In these cases, when hydrogen-induced cracking occurs, it is found predominantly in the weld metal (Coe 1973). Often, hydrogen-induced cracking, which originates in the weld metal, propagates into the heat-affected zone. When welding parameters are altered to alleviate hydrogen-induced cracking, the cracking in the heat-affected zone disappears first.

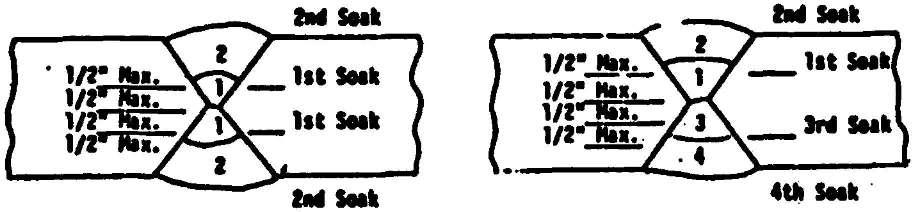
TABLE 6-1 MIL-14018 Weld Metal Hydrogen Removal Requirements

Category	Weld thickness (see 13.9.1) (in)	Repair depth (in)	Minimum excavation soak		Minimum intra-weld[2/][3/] soak (see figures below)		Minimum post weld soak[2/]	
			Temp (°F)	Time (hr.)	Temp (°F)	Time (hr.)	Temp (°F)	Time (hr.)
Original welds	1/2 and less	NA[1/]	NA	NA	NA	NA	250	4
	Greater than 1/2	NA	NA	NA	250	10	350	12
Weld repairs for cracks only	1/2 and less	All	250	4	NA	NA	250	4
	Greater than 1/2	1/2 & less	250	4	NA	NA	350	8
		Greater than 1/2	350	6	250	10	350	12
Weld repairs for other than cracks	All	1/2 & less	None	None	NA	NA	250	4
		Greater than 1/2	None	None	250	10	350	12
Interruption of preheat (see 13.7.6.3)	1/2 and less	NA	NA	NA	NA	NA	250	4
	Greater than 1/2	NA	NA	NA	NA	NA	350	12

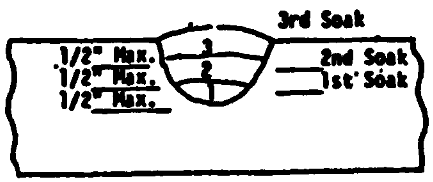
1/ NA - Not applicable.

2/ If intra-weld soak is performed, a postweld soak is not required. If any of the intra-weld soaks are not performed a postweld soak is mandatory.

3/ When intra-weld soaking is employed, a soaking treatment shall be performed on each 1/2 inch or less of weld deposited on any one side of the joint.



Acceptable Intra-Weld Soak
Sequence for Original Weld



Acceptable Intra-Weld Soak
Sequence for Repair Weld

Source: David W. Taylor Naval Ship R&D Center 1978.

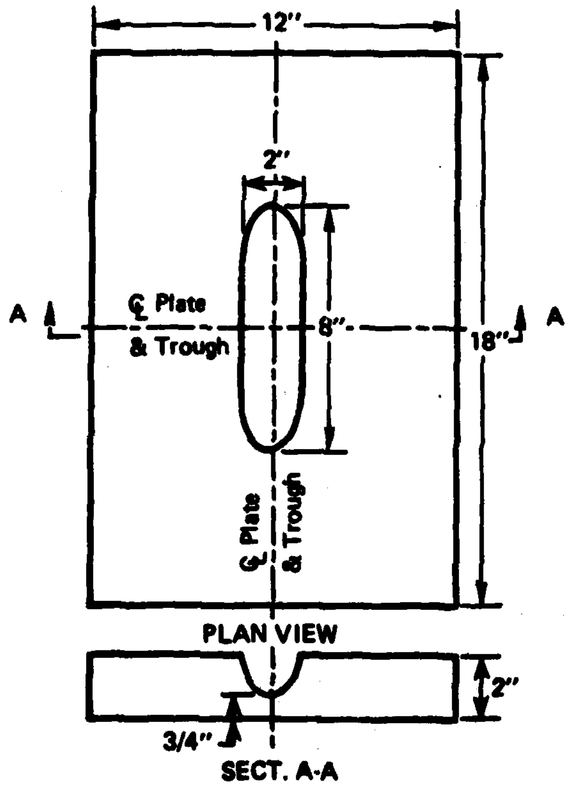


Figure 6-13 DTNSRDC tough Weldability specimen.

Source: Juers 1975.

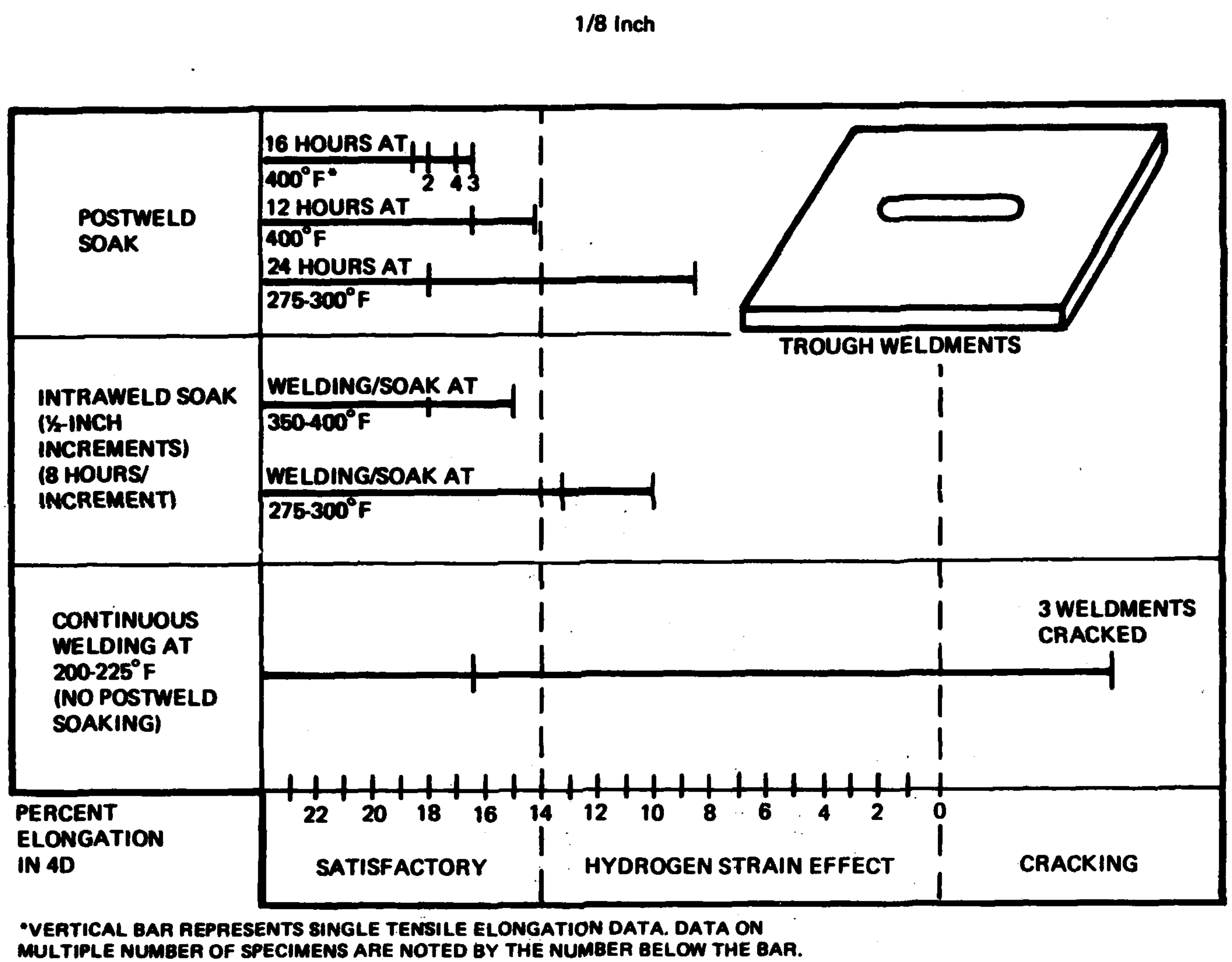


Figure 6-14 McKay 14018 laboratory trough weldments made with 1/8 inch diameter electrode.

Source: Juers 1975.

Conclusions

1. Hydrogen-induced cracking, a time-dependent fracture, may occur in the heat-affected zone and the weld metal if four conditions are present simultaneously: a critical stress level, a critical diffusible hydrogen concentration, a susceptible microstructure, and a temperature in the critical range of -150°F to 400°F.

2. In the welding of HY-steels, essentially nothing can be done to change the susceptibility of the microstructure. The welding parameters are chosen so as to produce a tough self-tempered martensite structure in the HAZ. This structure is susceptible to hydrogen-induced cracking.

3. Nothing can be done to change the HAZ operational temperatures that fall essentially in the middle of the critical temperature range of susceptibility, -150° to 400°F.

4. The susceptibility of steels (including HY-steels) to hydrogen-induced cracking increases with increasing yield strength of the weld and the heat-affected zone.

Recommendations

The following recommended actions should reduce the susceptibility of HY-steels to hydrogen-induced cracking:

1. Weld with materials and in environments that minimize hydrogen production, particularly low-moisture content conditions in the case of SMAW.

2. Use welding systems that produce weld metals with a minimum acceptable strength.

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CHAPTER 7

WELD PROPERTIES

The service performance of ferritic weldments depends in part on the properties of three distinct elements of the metallurgical system: the weld metal, the prime plate and the heat-affected zone of the plate. Performance also depends on whether these individual elements are stressed statically, in fatigue, or in impact. Additionally, performance is affected by stress concentrations from joint design, discontinuities at the weld reinforcement, and subsurface microfissures associated with inclusions, hydrogen-induced cracks, or corrosion. The weld metal plays a dominant role in establishing the properties of welded joints. This is so not only because the weld metal may be the weak link in the system, but for several additional reasons: the thermal energy introduced by superheated weld metal alters the plate's properties by producing the heat-affected zone; the hydrogen released from the weld metal is a potential source of defects; the weld reinforcement can intensify stress concentrations in the coarse-grained region of the plate's heat-affected zone; and defects resulting from poor welding procedures can nucleate premature failure.

Weld Chemical Composition

The composition of weld metal affects fabrication costs significantly. Higher quality is not obtained without more care, more expensive materials, lower deposition rates, and, therefore, higher costs. A discussion of the effect of weld metal on performance must consider not only its composition, but welding processes and welding procedures as well.

Microstructural Requirements

Iron can be alloyed with a number of elements to produce steels having vastly better properties, particularly higher strength. The strengthening mechanisms involve solid-solution hardening, treatments in conjunction with controlled alloying to introduce other phases such as carbides, and allotropic transformations. The filler metals used to fabricate the steels of primary concern in this report (strengths ranging from 80 to 130 ksi) are alloyed to satisfy a number of requirements which may appear to be contradictory. The filler metals must be chemically compatible with the base plate, since some alloying with it is inevitable. They must produce an unusual combination of high strength and toughness in the as-welded condition, and at solidification and cooling rates that are consistent with proven welding processes used to fabricate steels ranging in thickness from fractions of an inch to heavy plate. And they must be relatively inexpensive.

These requirements have been shown to be achievable with low-carbon, acicular microstructures called bainites and martensites (Figure 7-1). Briefly, the most suitable weld metals contain less that 0.12 percent carbon; they obtain the necessary hardenability and some solid-solution strengthening from their content of manganese, nickel, chromium, and molybdenum; they are deoxidized with silicon and titanium; and the heats are carefully controlled to minimize the presence of potentially embrittling elements such as vanadium and tin.

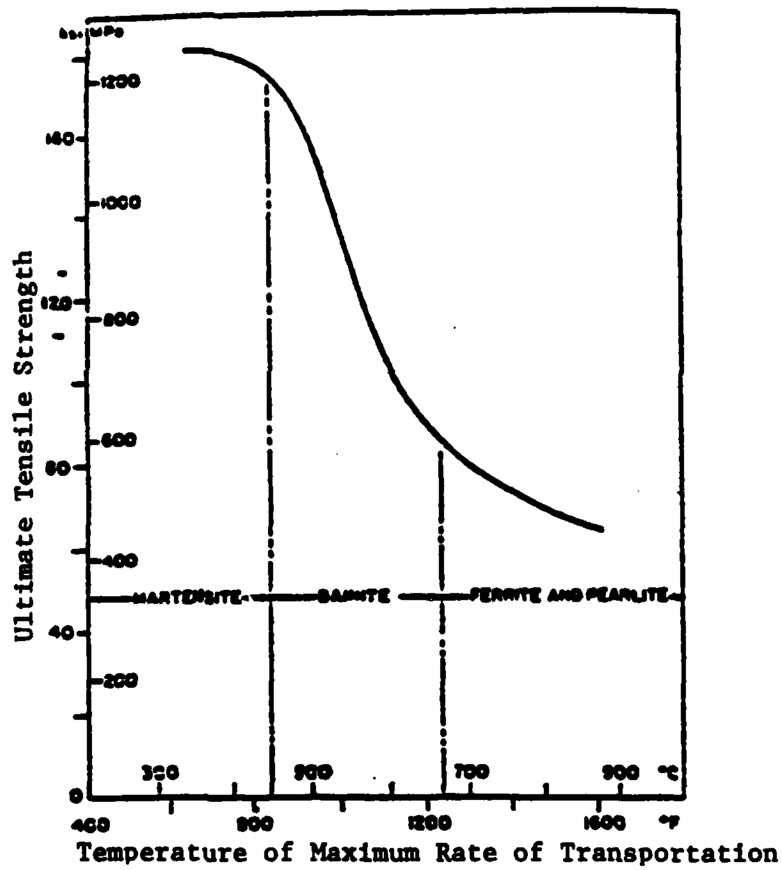

Figure 7-1 Effect of transformation temperature on strength.

Source: Irving and Pickering 1958.

The properties required in these weld metals are obtained by proper alloying and by controlling the cooling rates through the subcritical temperatures at which the austenite transforms. Lower-strength alloys generally transform to bainite (Figure 7-2) and higher-strength alloys to martensite (Figure 7-3). Obviously, the cooling rates used must be within the ranges expected when fabricating steels with commonly used arc welding processes. With lower cooling rates, the transformation products may not be acicular and, in any event, would be weaker than desired. Because of problems associated with coarse grains in the heat-affected zones of the steels to be welded, the welding alloys were designed to produce acceptable microstructures at relatively low energy inputs--below 45,000 Joules per inch and with preheat temperatures not exceeding 350°F (180°C).

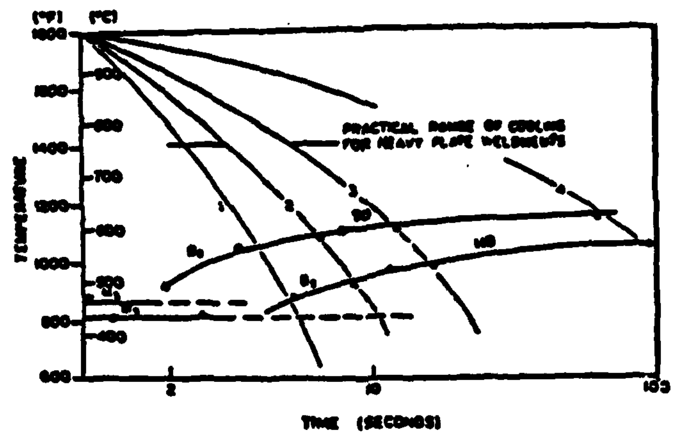

Figure 7-2 CCT diagrams for the 90 and 110 ksi yield strength weld-metal systems.

Source: Lyttle et al. 1969.

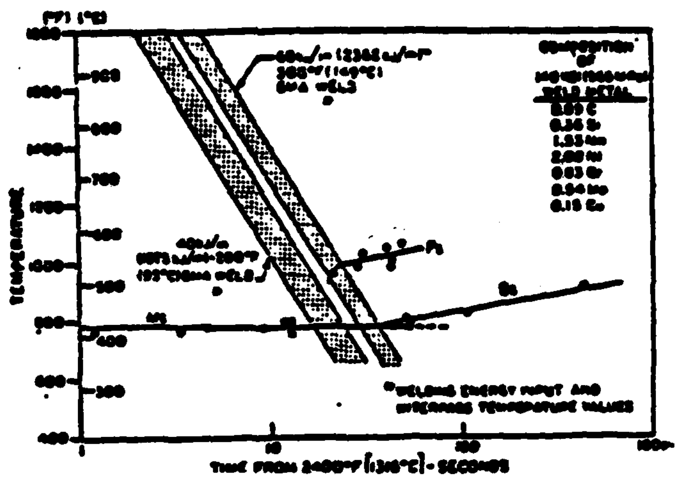

Figure 7-3 Continuous cooling transformation diagram for 140 ksi yield strength weld metal.

Source: Krantz 1971.

Alloying Effects on Strength and Toughness

The low-carbon, acicular microstructures are unique because they are capable of being strengthened without losing toughness or ductility. Because welding techniques affect cooling rates and weld characteristics, practical welding procedures were anticipated and used during the development of these alloys. For example, the oxygen needed to stabilize argon-shielded metal inert gas (MIG) arcs will reduce the heat transfer efficiency of the more easily oxidized alloys such as those containing manganese and chromium. Other undesirable effects may occur in other welding environments such as the active regions of covered-electrode arcs.

As stated earlier, weld metal alloys must be carefully matched with the welding systems to produce the necessary strengths without impairing toughness. Very briefly, the main factors are as follows: the manganese and nickel levels can be increased to about 2 percent to reach the greatest

strength, while simultaneously improving toughness; subsequent increases in strength require the addition of chromium and carbon, but only enough to provide the hardenability and carbides necessary; molybdenum is commonly used to shift the bainite nose (in the phase diagram) and avoid the formation of high temperature transformation products.

Inclusions

Stronger ferritic weld metals are less tolerant of the embrittling effects of inclusions such as sulfides, oxides, and nitrides. For example, toughness can be improved dramatically by reducing sulfur to below 0.015 percent (Figure 7-4) or by removing sources of oxides from the welding system (Figure 7-5). For this reason, tungsten inert gas (TIG) welds made in oxygen-free argon are tougher than MIG welds in oxygen-doped argon; both are tougher than welds with similar compositions and strengths deposited with covered electrodes. Nitrogen contamination also must be avoided (Figure 7-6).

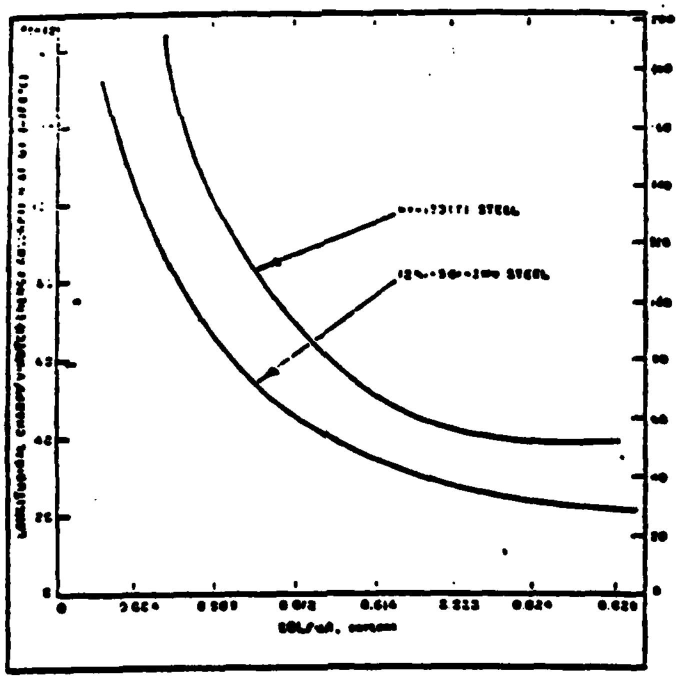

Figure 7-4 Effect of sulfur on the notch toughness of HY-130 (T) and 12Ni-5CrMo steels.

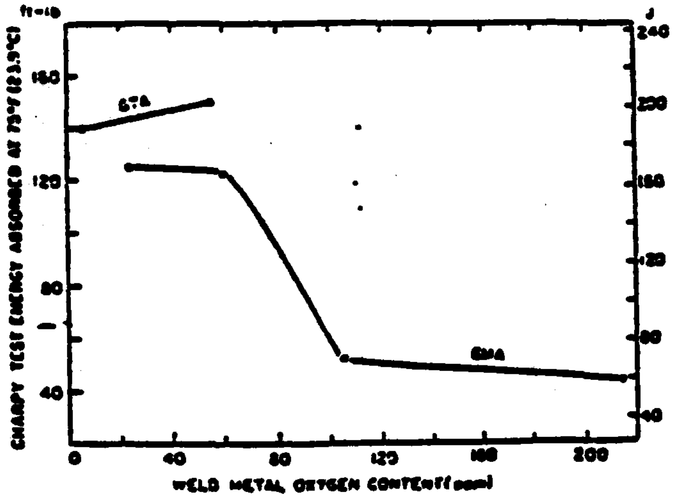

Figure 7-5 Effect of oxygen content on Charpy V-notch shelf energy of 3% weld metals.

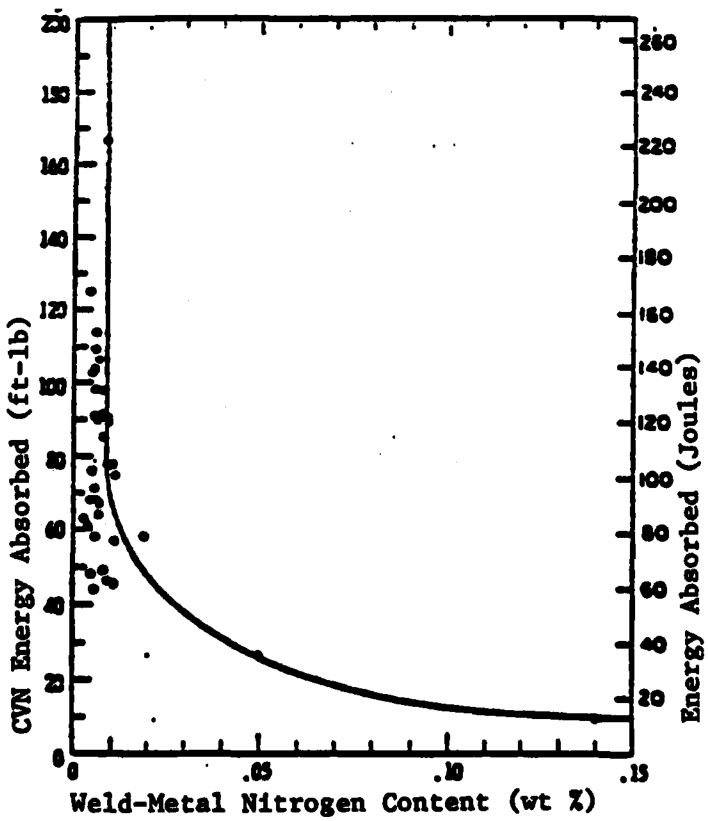

Figure 7-6 Nitrogen dependence of Charpy V-notch impact values at 80°F (26.7°C) for unalloyed carbon steel weld metals (all processes).

Because the effect of inclusions on toughness is more significant with higher-strength steels, more care must be taken to produce cleaner welds with such steels. One approach is to take care in deoxidizing the steels to be converted to filler metals. Enough silicon is added to prevent porosity. Titanium also has been used in filler metal as a scavenger for the oxygen used in MIG welding (Figure 7-7); the effect on the toughness of the weld metal can be dramatic, but the amount of titanium needed is highly dependent on the oxidizing potential of the shield gas.

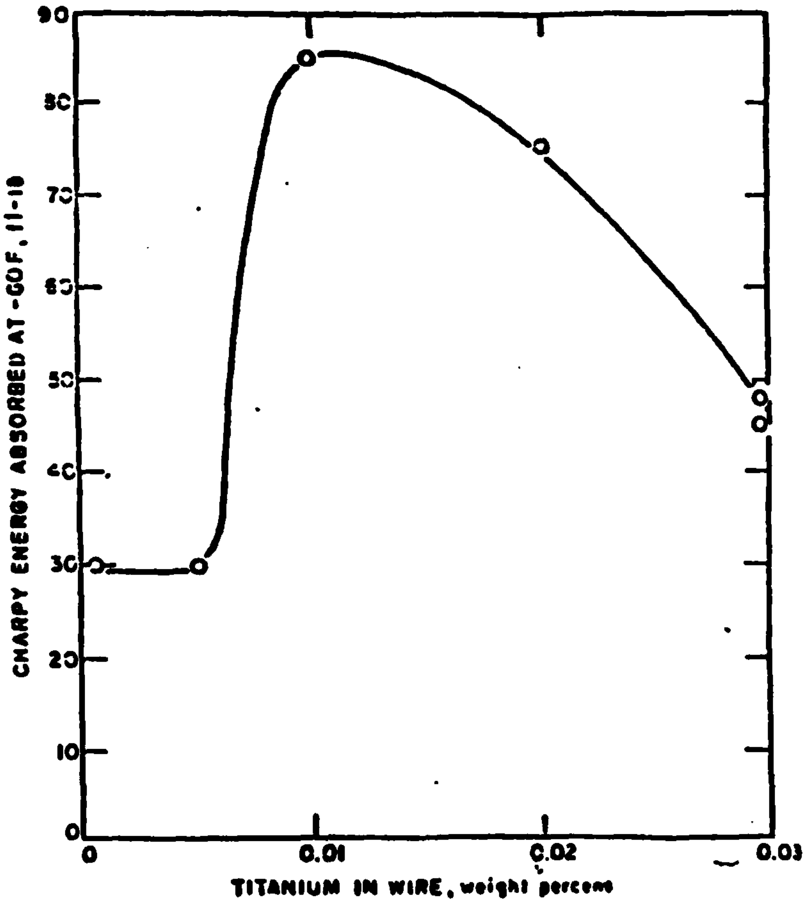

Figure 7-7 The effect of titanium in filler metal on toughness of weld metal.

These observations reemphasize the need to tailor systems by matching the shield-gas composition, the filler-metal composition, and welding procedures to obtain a clean and properly deoxidized weld metal that will transform to the most suitable microstructure after welding.

Inclusions can also be controlled by incorporating basic fluxes with processes such as covered electrodes, submerged arc, or cored wires. Unfortunately, increasing the basicity also tends to degrade the operational characteristics of flux-containing welding processes. With cored wires, for example, the basic flux systems are characterized by spattery, globular transfer and fluid, adherent slags. They are thus very difficult to use in the vertical position.

The EXX18 class of covered electrodes are basic, but can be used with relative ease for depositing satisfactory welds in all positions (Figure 7-8). However, if care is not taken, air can be entrained in the arc when welding vertically, allowing enough nitrides to precipitate in the weld metal to reduce notch toughness.

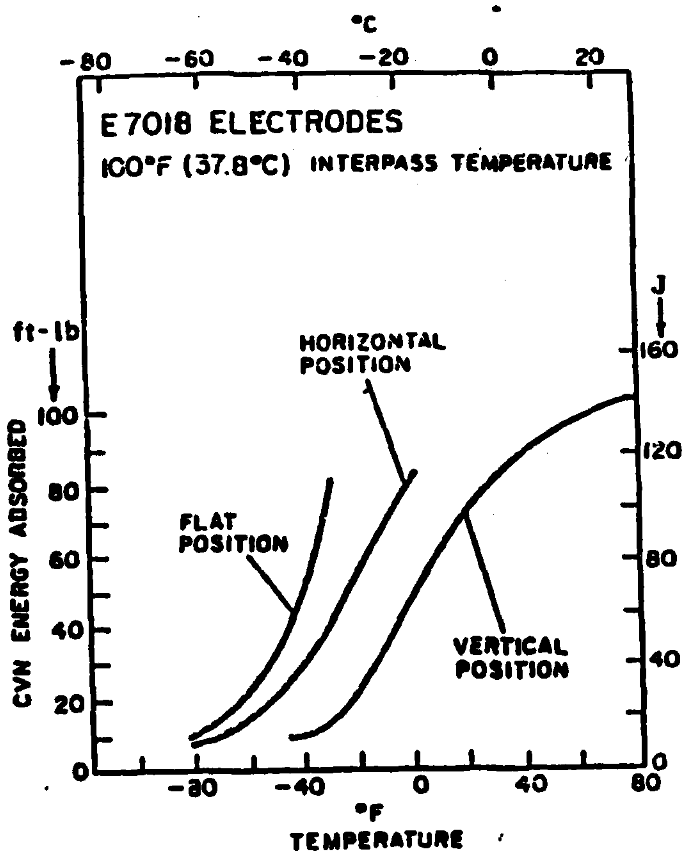

Figure 7-8 The notch toughness degradation observed in some out-of-position mild steel weldments.

Hydrogen

As described in Chapter 6, hydrogen induced cracking is one of the most aggravating problems associated with welding HY-steels. Hydrogen comes from the residual moisture and organic materials associated with some constituents of electrode coatings and submerged arc fluxes. Hydrogen also is present on the surface of MIG wires, in lubricants and corrosion inhibitors.

The damage produced by hydrogen appears to be reduced if not eliminated using techniques designed to reduce either the stress or the amount of diffusible hydrogen, or both. Preheating has been found to be beneficial probably because it reduces the cooling rate at a given energy input, which results in a lower hydrogen content.

Higher welding energy inputs might be used to allow hydrogen to diffuse from the weld in the same manner as produced by preheat. However (as will be explained in detail in a later section), the energy input increases that are required to reduce cooling rates at temperatures approaching 400°F would be exorbitant. This is because the cooling rates through the critical temperature region would be reduced enough to significantly degrade the metallurgical and mechanical properties. Of course, electrodes can be manufactured with lower hydrogen content to minimize the problem, if not eliminate it. Regardless of the approach taken to minimize HIC--the need for additional equipment and steps for preheat or for more expensive, lower-hydrogen electrodes--fabrication costs will increase in direct proportion to weldment strength (Boniszewiski and Moreton 1969).

Inclusions such as sulfides may act as sinks for diffusible hydrogen, thereby reducing the amount of preheat required to prevent HIC (Hirai, Minakawa and Tsuboi 1980). Ordinarily, this may not be relevant because the sulfur levels are above those shown to be potentially harmful. However, if sulfur must be reduced to 0.005 percent to obtain adequate toughness, the preheat temperature may need to be raised by 150°F to 175°F (65°C to 80°C) above that shown to be adequate for the same alloy with normal sulfur.

Summary

In summary, the problems of fabricating steel weldments increase in proportion to their strength. As strength increases, more care is needed to manufacture properly alloyed and hydrogen-free electrodes, and more attention is needed to control preheat and arc-energy inputs. The additional controls and effort increase the overall costs of large structures and increase the risk of poor quality. Therefore, in the interest of cost-effectiveness and reliability, the lower-strength alloys are favored and should be used when possible.

Potential Solutions to the HIC Problem

A number of techniques have been used to minimize if not eliminate hydrogen-induced cracking. They involve proper alloying to reduce the weld strength, introducing preheat or increasing the arc energy to reduce the cooling rate, reducing the hydrogen associated with filler metal, or modifying the joint design to reduce residual stresses. The simplest approach involves low-hydrogen filler metals with tensile strengths close to those of the plates being welded.

Low-Carbon Alloys

Experience has shown that HIC can be minimized by keeping the carbon content below 0.10 percent. Although this can be done easily with weld metals designed to produce yield-strengths below 100 ksi, more carbon is necessary to obtain high strength in thick, quenched and tempered plate, or in the weld metal.

Low-Carbon Equivalent Alloys

Steels with greater hardenability (carbon equivalence) produce microstructures that are more sensitive to hydrogen. Modification of alloys to maintain the lowest possible carbon equivalence, therefore, would be expected to minimize the incidence of HIC. Unfortunately, this approach is doomed to failure because metallurgists have not been able to develop high-strength steels having low hardenability. Alloying is essential to produce appropriate microstructures, and richer alloys are needed to meet requirements for greater strength and thicker structural sections. The alloying of these steels is complicated further by the need for high toughness.

Preheat

At one time, the panacea for HIC was to "use XX18 electrodes and preheat." Preheat reduces the rate at which welds cool and provides a relatively high base temperature to accelerate the escape of hydrogen from the system before the critical nucleation time for HIC is reached. Unfortunately, the cooling rates of high-strength steels must be controlled to obtain the required microstructure, either bainitic at strengths close to 80 ksi, or martensitic at strengths approaching 130 ksi. Excessive preheat will allow carbides to agglomerate, significantly reducing strength. However, if the preheat temperature is increased in proportion to plate thickness, it is possible to maintain optimum cooling rates through the critical temperatures at which transformation occurs. As a matter of fact, by preventing excessively high cooling rates, potentially sensitive microstructures can be avoided.

Although a number of empirical results have been used to demonstrate that preheat prevents delayed cracks, the explanations may not be valid. All rely on accelerating the escape of diffusible hydrogen so that not enough hydrogen atoms can concentrate at nucleation sites tc produce cracks. However, hydrogen diffusion rates would be higher with preheat and HIC might occur more quickly. More work is needed to explain adequately the beneficial effects of preheat on HIC and to learn whether other, less expensive techniques could produce equivalent results. For example, the results obtained with preheat might be explained metallurgically.

Arc-Energy Input

Another way to combat HIC is to reduce the cooling rates of welds by increasing the arc energy. This approach may not be desirable, however, because the size of the weld pool also is increased. Massive welds are especially more difficult to control when made in the vertical position. They produce a larger heat-affected zone which can reduce both the strength and toughness of welds (Figure 7-9). These effects are similar to those produced by preheating. Therefore, arc-energy input cannot be considered independently of the preheat temperature.

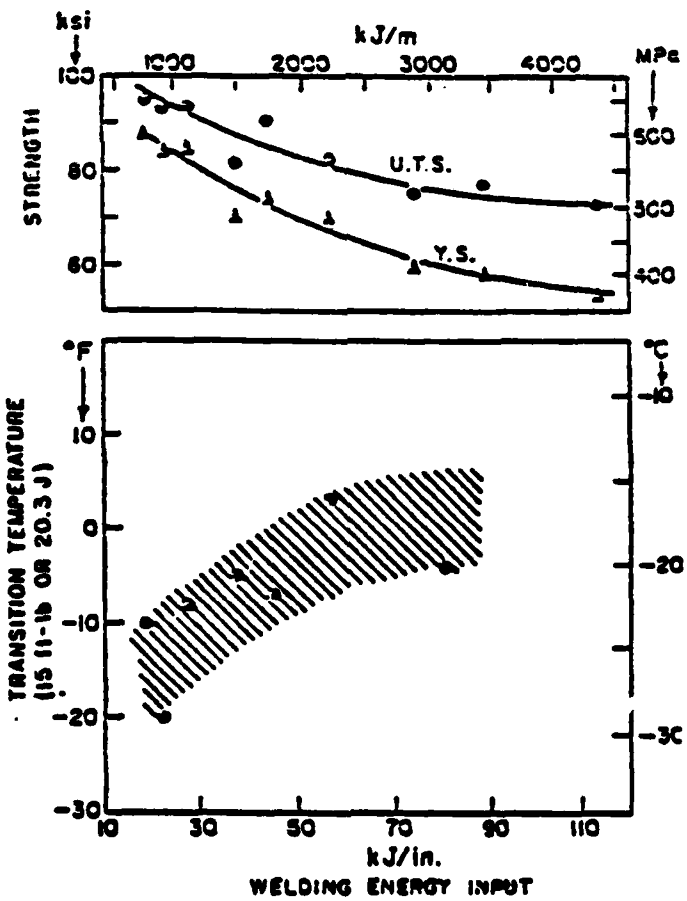

Figure 7-9 Effect of energy input on notch toughness and strength of gas-shielded arc weld metal.

Weld Cooling Rates

The interaction of energy input and preheat temperature on the cooling rate of welds can be described quantitatively by the expression

$$\frac{dT}{dt} = M \frac{(T - T_o)^2}{E} + D.$$

In this expression, the cooling rate dT/dt is that found at any temperature T during the cooling cycle from the solidification temperature to the final temperature of the weldment, T_o, which is then the preheat or interpass temperature. E is the energy input per inch of weld, calculated from the voltage, current and travel speed. The constants M and D are determined experimentally. This expression is valid only for three dimensional cooling situations; i.e., those involving plate thicker than one inch.

The plate thickness has an effect, particularly when the plate is so thin that it extracts heat in two dimensions. In heavier sections where three-dimensional cooling is possible, welds cool significantly faster. The effects of plate thickness, preheat, and welding energy input on weld-metal strength are summarized in Figure 7-10. Since the strongest microstructures

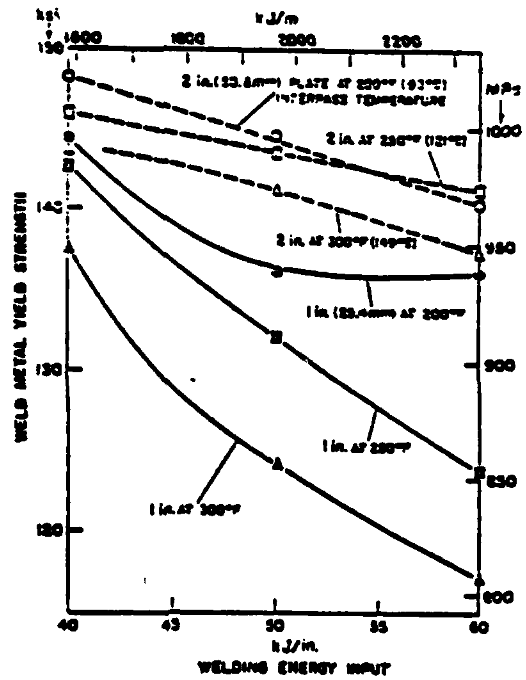

Figure 7-10 Tests with 1 and 2 inches steel plates show that the use of low heat inouts and low interpass temperatures produce the strongest welds.

are associated with high transformation rates (Figure 7-1), reducing cooling rates should significantly affect strength. The rates are highest with heavy sections, low preheat temperatures, and low welding-energy inputs. The rates decline with thinner plates, higher preheat temperatures, and at higher welding energies. The strongest welds in Figure 7-10 (148 ksi) were obtained in 2 in. plate with preheat, T_0, of 200°F (93°C) and a relatively low energy input of 40 kilojoules per inch. Reducing the plate thickness to 1 in., raising the preheat temperature of T_0 to 300°F (149°C) and increasing the energy input to 60 kilojoules per inch reduced the strength to 118 ksi.

By selecting various combinations of these three variables, it is possible to change the cooling rate and thus the strength of the weld metal. The data of Figure 7-10 are plotted according to cooling rate in Figure 7-11; confirming data are shown in Figure 7-12. Obviously, welding conditions must be selected to prevent significant softening of weld metal caused by its transformation at high temperature to an undesirable microstructure. Preheat may effectively deter HIC by reducing cooling rates to allow transformation to somewhat weaker microstructures that are more tolerant of hydrogen. However, if excessively high preheat temperatures are used, the retarded cooling rates could weaken the weld enough to impair the performance of the welded structure.

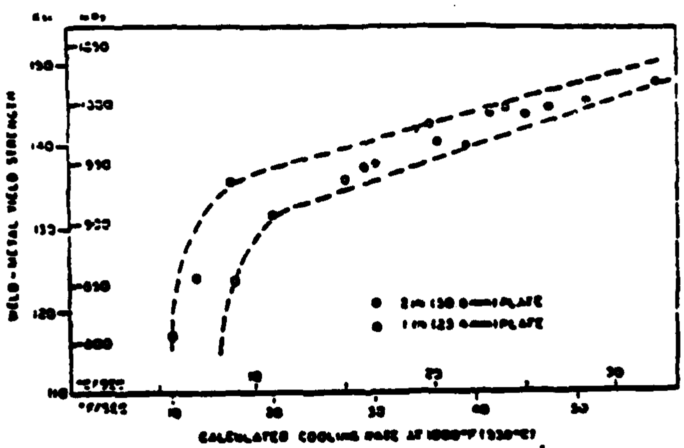

Figure 7-11 The effect of weld metal cooling rate on yield strength.

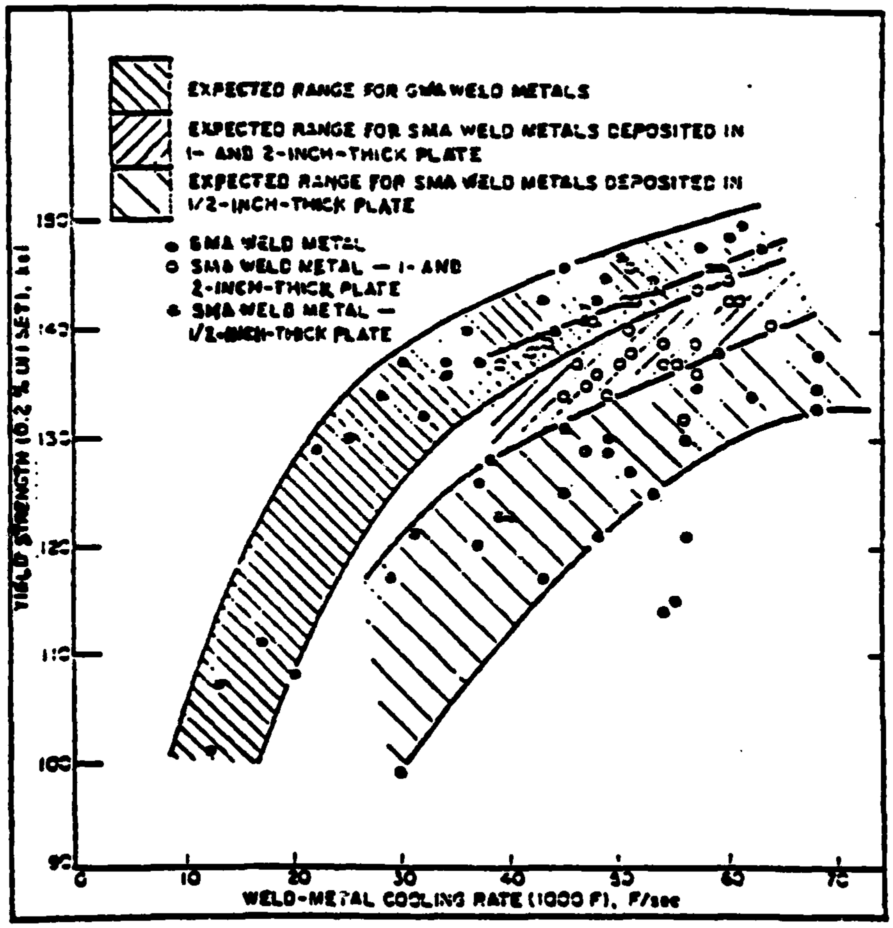

Figure 7-12 Relation between yield strength and weld-metal cooling rate for HY-130 (T) weld metals.

The real effect of preheat is sensed more at low temperatures; lower than those at which austenite is transformed. For example, with a typical energy input of 50,000 Joules per inch, increasing the preheat T_0 from 72°F to 250°F retards the cooling rate by 50 percent at T of 1000°F and by almost 400 percent at T of 400°F. The net effect is that the mechanical properties would not be changed significantly, but the time allowed for hydrogen to escape would be considerably longer. Energy input could not be increased enough to obtain the same beneficial effect of reducing hydrogen without both softening the weld metal and extending the heat-affected zone. The energy input must to be increased from 50,000 J/in. to 240,000 J/in. to get the same 400 percent reduction cooling rate at 400°F as obtained by preheating to 250°F. This would reduce the cooling rate at 1000°F by about 400 percent, and reduce the strength of the weld significantly.

Low-Hydrogen Filler Metals

In the long term, the most cost-effective solution to the HIC problem may be to control hydrogen at its sources: the electrode, associated fluxes or shield gases, and contaminants on the plate. Hydrogen levels associated with processes using solid wires shielded with inert gases can be significantly lower than those requiring fluxes for protection. The fluxes may contain hydrogen-rich constituents. More likely, they can absorb significant amounts of moisture from the air unless care is taken in their manufacture, packaging, and storage. Frequently overlooked sources of hydrogen are contaminants in the joint to be welded. Paints, oils and greases, organic debris, and rust are typical sources of hydrogen that can cause HIC and porosity. Later sections in this review discuss the relative levels of hydrogen associated with different welding processes.

Welding Processes

The welding processes in this section of the report are limited to those proven to be cost-effective for assembling steel structures in large fabrication shops. This limitation implies such conditions as imperfect fit-up, commercial tolerances in thickness, welders of average skills, and limited supervision. The tungsten arc process, although used successfully in the aircraft industry and capable of depositing very clean weld metal, does not lend itself to fabricating large structures because the maximum deposition rates are much too low to be practicable. Other processes such as electron beam (EBW) and even narrow-gap (GMAW) require extremely precise fit-up to be considered practical. Therefore, only the merits and problems associated with the gas-shielded metal arc welding, cored wire, covered electrode, and submerged arc will be discussed.

Gas-Shielded Metal Arc

The gas-shielded metal arc process has a number of advantages for joining high strength steels. Because solid wires are used for filler metals, hydrogen levels can be kept very low. Even if the wires are coated to provide lubricity and rust resistance, hydrogen will not exceed 8 ppm. With special cleaning, it can be held to a total below 2 ppm, a level found necessary for joining HY-130 with no chance of developing hydrogen-induced cracks (see Figure 6-12). These very clean wires may require special packaging to prevent rusting, and feeding them may be troublesome unless care is taken with the design and maintenance of the welding equipment.

The spray arc mode is most commonly used because the argon-oxygen or argon-carbon dioxide mixtures, required for spray transfer, are relatively inert (commonly 2 percent oxygen or 5 percent carbon dioxide is used). The filler-metal compositions are designed for that level of oxidation so that sound and well penetrated welds are easily achieved, the deposition rates are moderately high, and the deposits have a good reinforcement profile and are relatively free of slag. With argon-rich gas shields the process is easy to control and relatively simple to use. Carbon dioxide shields are not recommended for welding high-strength steels because they are highly oxidizing as well as producing spattering arcs and highly crowned deposits. Without modifications, the spray arc mode is acceptable only for depositing welds in the flat or horizontal position. The arc is too forceful and the weld metal too fluid to be effective in the vertical position.

Pulse-arc power supplies were developed to reduce the average current and deposition rate while retaining the desirable characteristics of the argon-shielded spray arc process. The welder must be more skillful because the power supplies are more complex and the weld pool is somewhat more difficult to control. Undercut welds can be troublesome if welders are not skilled or the joints are not free of rust or mill scale. Fatigue can reduce the effectiveness of welders, limiting the amount of weld metal deposited during a shift. Some of these problems have been eased if not eliminated by mechanizing the pulsed-arc process for welding in all positions. By mounting the equipment on tracks and using oscillators to propel and manipulate the arc, high quality welds have been deposited in all positions, at reasonable deposition rates, and without the complications produced when the arc must be interrupted because of welder fatigue.

Another modification of the MIG process for all-position welding is the short-circuit arc. This technique produces a stable, high-frequency short circuiting of the molten wire tip and weld pool to transfer weld metal without spatter and with good control. It requires a matched system consisting of a specially designed power supply, small-diameter filler wires, and shield gases with a high voltage gradient. The arc energy is relatively low, allowing controllable deposition rates in all positions. With proper shield-gas mixtures, such as 75 percent argon and 25 percent carbon dioxide, very acceptable welds can be produced. Relatively little training is needed to develop proficient welders.

Unfortunately, the energy levels of the short circuit arc are insufficient to insure good fusion in heavier sections. Lack-of-fusion cannot be avoided when plate thicknesses exceed about 3/8 inch; something that limits the process essentially to fabricating sheet metal. Therefore, when radiographyically-confirmed integrity is required, welds made in heavy sections with short-circuit transfer may not be acceptable. Special ternary shield gases have been developed to improve penetration without impairing the desirable characteristics of the process. Even they have not been accepted generally for welding large sections because the need for weld repair has been reported to be excessive.

Gas-Shielded Flux Core

Two families of flux-cored wires are in common use for fabricating steels with gas-shielded processes. One is based on rutile-rich systems. The operational characteristics of this class are excellent. However, the resultant welds are too rich in oxides to produce the toughness required for fabricating structures intended for severe conditions. Another family of fluxes is based on fluoride systems. The welds produced with them are tougher and relatively free of hydrogen. Unfortunately, the operational characteristics of these electrodes are poor. Spatter, fume, slag removal, and bead profiles are troublesome. Also, these flux systems produce slags that are too fluid for depositing welds in the vertical position.

The rutile-type flux-cored wires have been used with great success for depositing welds in vertical positions at rates exceeding 7 lbs per hour. Using wires in diameters of 1/16 in. and smaller with argon-rich gas mixtures containing carbon dioxide, sound, spatter-free welds can be made by relatively unskilled welders. The welds have excellent penetration and good reinforcement profiles and are free from the undercut difficulties that may occur with the spray-arc process. Although cored wires are acceptable for joining mild steels, the toughness obtained with them in high-strength steels cannot meet the levels specified for severe service.

Cored wires are not restricted to all-position applications. Originally they were designed to produce welds in the flat and horizontal positions at deposition rates exceeding 20 lbs per hour. Whether high-strength steels can tolerate the energy needed at such high deposition rates is questionable. However, recent experiments with submerged arc weldments have shown that higher energy inputs may be used in some applications without impairing the performance of heavy-walled structures.

Covered Electrodes

Covered electrodes were developed before World War I and remained the only arc welding process enjoying significant acceptance until a few years after World War II. Since there was little competition from other processes, much time and effort were given to developing a variety of flux systems that

provided great versatility for the industry. Shortly after hydrogen was recognized as the cause of delayed cracking, the low-hydrogen electrodes incorporating carbonates and fluorides in the coating were developed from those used for stainless steels. The existing XX18 class evolved as a result of the addition of iron powder to improve deposition efficiency and operational characteristics. Recent modifications led to the introduction of coatings having the virtues of low hydrogen levels in the as-baked condition and low rates of moisture pickup when exposed to hot, humid air (Figures 7-13 and 7-14).

Covered electrodes have survived the competition of new, more efficient processes because they can be used to deposit acceptable welds in all positions, in confining locations, and without the encumbrances of hoses, feeders, and electronic controls. At the present state of evolution of welding processes, covered electrodes alone have the general characteristics required of a universally acceptable tool. Their faults are primarily economic; the electrodes are not efficient, have low deposition rates in the vertical or overhead positions, and are reliable only when used by properly trained welders.

Until recently, moisture levels in covered electrodes could not be maintained reliably at the 0.2 percent or 0.1 percent ceilings needed, respectively, to weld the HY-80 and HY-100 classes of steels. Complete assurance of crack-free welds is possible for HY-130 weldments if moisture can be kept below 0.05 percent (Figure 6-3). To be practical, the electrodes should not absorb moisture when exposed to the humid environments associated with shipyards. But this requirement is not met easily, and electrodes produced by some manufacturers absorbed moisture at notorious rates (Figure 7-15). This characteristic brought the additional economic burden of higher preheat temperatures. It is doubtful that the HY-130 steels can be welded satisfactorily using covered electrodes without adequate preheat. However, this requirement might be eliminated with low-moisture electrodes or fabricating the leaner, more tolerant HY-80 steels.

This possibility is given more credence by Peng at Ohio State (1981) who showed that a "lower critical stress" of about 91,000 psi could be obtained with conventional E11018 electrodes merely by preheating to 250°F. With low moisture electrodes, however, the "lower critical stress" of 88,000 psi was reached without any preheat. The diffusible hydrogen levels of welds made with the moisture resistant electrode were 1.00 ml/100g, compared with 1.92 for the conventional electrodes. The "lower critical stress" is that level which can be sustained indefinitely without failure when using an implant test. Without preheat, the lower critical stress with the conventional E11018 electrode was only 59,000 psi.

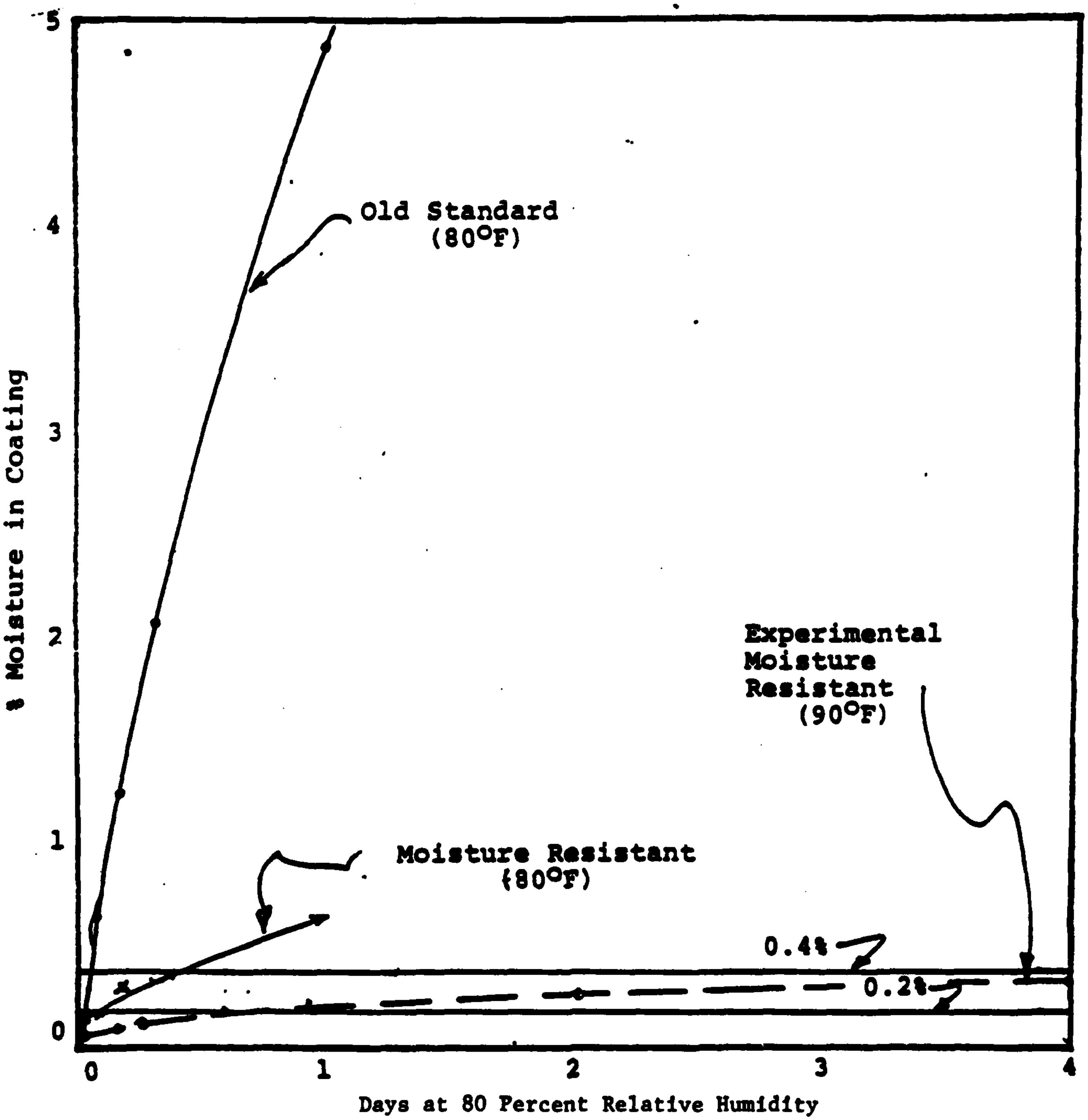

Figure 7-13 Improvements achieved with moisture resistance of XX18 covered electrodes.

Source: AIRCO internal research data.

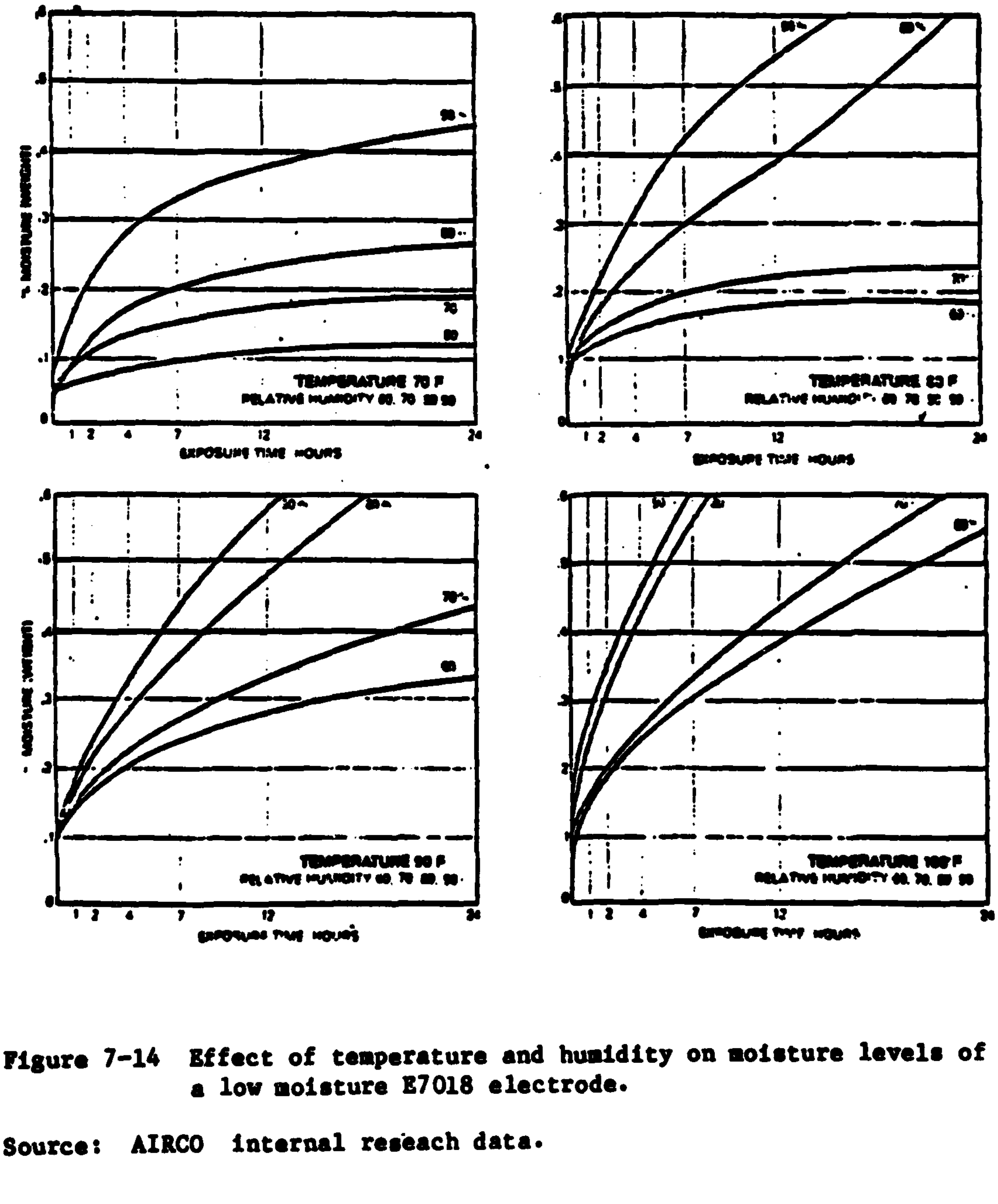

Figure 7-14 Effect of temperature and humidity on moisture levels of a low moisture E7018 electrode.

Source: AIRCO internal reseach data.

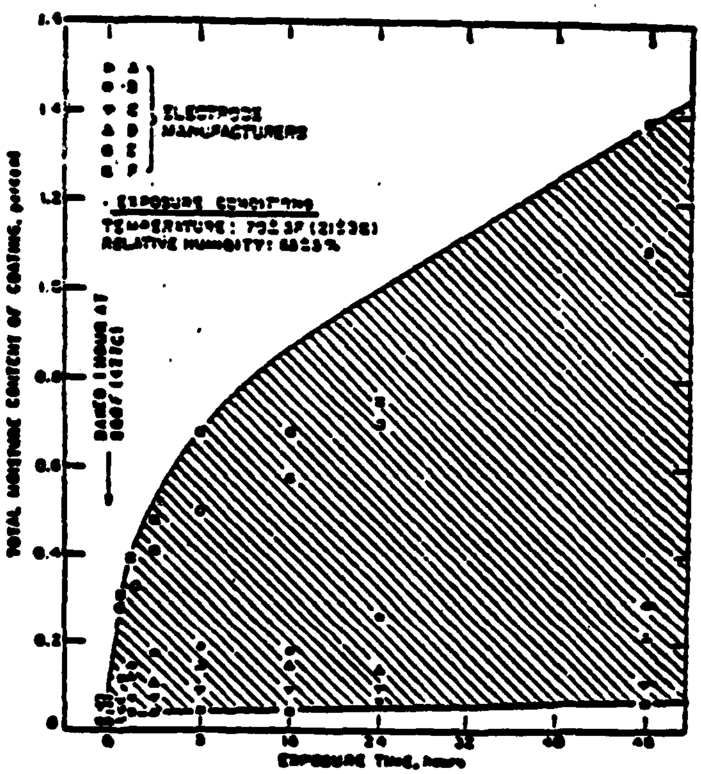

Figure 7-15 Moisture content of AWSE7018G electrodes exposed to an atmosphere of 65% relative humidity (49).

Source: Wilkbrand & Gilkison 1973.

Submerged Arc

The submerged arc process for joining high-strength steels has had a mixed reception from welding engineers. The process is capable of depositing sound weld metal at high deposition rates, and penetration can be very deep, which simplifies joint design and significantly reduces machining costs. Little welder training is needed. However, welds must be made under a blanket of flux that obscures the joint, and they can be made only in the flat or horizontal position. The welds must be properly positioned with respect to the welding head, necessitating the use of fixtures and mechanized equipment. The flux can abrade gears and jam bearings.

These disadvantages are tolerable when full advantage can be taken of the unusually high deposition rates possible with the submerged arc. Unfortunately, the requirement for controlled cooling rates has severely restricted the maximum acceptable energy of 50,000 J/in. input compared with

about 100,000 J/in. for cost-effective submerged arc welds. This circumstance favored the use of other processes, the gas-shielded processes in particular. Although less efficient than the submerged arc, they were favored when compared at the same relatively low energy inputs specified for welding high-strength steels. Additionally, the gas-shielded arc welds have been found to be tougher.

The most satisfactory fluxes for joining high-strength steels by the submerged arc process are neutral or basic. All alloying is provided by the filler wire. Basic fluxes are favored because they produce cleaner, tougher welds and are not as likely to produce hydrogen-induced cracks. Even so, they must be dried and properly stored to minimize the possibility of generating hydrogen.

Summary

No single welding process has all of the characteristics needed to ideally join high-strength steels. Although the MIG spray process is capable of depositing the toughest welds, and can provide reasonable deposition rates, it is not a good candidate for welding in the vertical position. This is done readily with the use of pulsed-power MIG process that reduces the average current input and permits lower deposition rates without losing the desirable spray transfer characteristics of the MIG process. Submerged arc welds can be more economical if advantage can be taken of the potentially high deposition rates in the flat position. Although submerged arc welds may have acceptable toughness, the MIG welds are superior. Good all-position welds can be obtained with small diameter, acid type cored wires. However, the toughness of such welds is barely acceptable even at lower strengths. Covered electrodes emerge as the most generally accepted method for all-position welding because their combined operational characteristics and weld properties are a uniquely acceptable compromise of versatility of application and final properties.

Weld Metal Deposition Costs

A number of factors contribute to the cost of fabricating heavy structures with high strength steels. The costs of filler metals, fluxes, and gases are only a minor part of the overall cost. The majority of the costs are related to the labor required for machining, positioning, and preheating the assembly, fabricating the weld, cleaning the weld between passes, and preparing the weldment for nondestructive testing and inspection.

The welding process affects the cost of each element of the welding procedure. It affects the precision of machining and the fit-up, the size of the joint cavity (amount of filler metal required), the rate at which weld metal is deposited, the ease with which it can be cleaned of slag and spatter, and the need for additional preparation such as preheat. The labor cost per pound for welding is easily calculated by dividing the hourly labor rate by the hourly weld metal deposition rate. Since welding processes can affect both the deposition rate and efficiency of the weld metal, they also affect the total time needed to deposit a given amount of metal.

The powerful effect of the welding process on the cost of labor needed to deposit weld metal is illustrated in Figure 3-1. The labor costs illustrated are in dollars per pound and are based on a purely arbitrary direct labor and overhead charge of $15 per hour. Obviously, the labor cost per pound of weld metal deposited will be reduced when a more rapid and efficient welding process is used. Also important, however, is the amount of time he spends in actual welding; his efficiency or duty cycle. When fabricating joints that are positioned vertically, a welder using covered electrodes generally spends about 15 percent of his time actually welding and deposits between 2 and 3 lbs per hour. Therefore, the cost of his labor is between $30 and $60 per lb of metal deposited as shown in Figure 3-1.

Changing to a process such as cored wires that permit a higher rate in the vertical position, a welder spends more time in actual welding. The labor costs thus can be reduced to $7 to $15 per lb of metal deposited. If the joint can be positioned so that the welds are deposited in the flat position, still higher deposition rates and efficiencies are possible, and the welder costs drop to the range of $1.50 to $3 per lb.

Obviously, the cost of the filler metal is important too. This depends on the cost of the electrode and the amount of electrode that can be converted to weld metal. For example, about 50 percent of a stick electrode is deposited as metal, so a fifty cent electrode deposits as a one dollar weld metal. Similarly, because of a much higher deposition efficiency, an equivalent sixty cent solid wire costs about sixty two cents as weld metal. The cost of a shielding gas would add about twenty cents to this weld metal cost.

Adding the costs for the filler metal, shielding gas, and welding labor, the total cost of welds made in HY-80 steels with various process is illustrated in Figure 3-2. The economic advantages of vertical welding using cored wires for manual welding, or using solid wires for mechanized welding, are evident when compared with covered electrodes. Costs can be reduced from about $45 per lb to $12 or $8 per lb, respectively. Although the difference is not as dramatic, the cost of SMAW is less than that of the MIG process; this assumes that the required setup times for each does not require significantly more time or reduce the welders duty cycle.

The industry also has a justifiable interest in avoiding costs that may be unnecessary, such as those for preheating massive sections in preparation for welding. The cost of labor and power have been shown to be significant. The delays in production caused by preheating are equally important and add substantially to overhead. More work should be done to define the precise role of preheat in minimizing hydrogen-induced cracks when welding the HY-steels of concern in this report. The tolerance of steels to hydrogen also needs to be reexamined. Since both variables affect HIC, their relationship needs very careful review. The sources of hydrogen in weld metals can be reduced by selecting low hydrogen processes and by modifying the electrodes. Such measures may reduce if not eliminate the need for preheating.

Also worth exploring is the influence of higher deposition rates on cost; since higher deposition rates require higher arc energies, their influence on the mechanical properties of the welded composite also needs to be considered (Figures 7-9 and 7-10). Savings in welding costs at the expense of weld properties certainly are not cost effective. Before undertaking such a study, however, it would be prudent to explore the real savings possible by substituting the submerged arc for the MIG process. The preparation time for aligning joints and the automatic equipment for submerged arc welding may be sufficiently greater to offset the higher deposition rates. Even with the same preparation time, doubling the deposition rate from 10 to 20 lbs per hour reduces the labor cost by 75 cents per lb. Compared to the possible labor savings of $35 per lb in the earlier example, this improvement does not appear quite as significant.

In reviewing the literature for this report, the most relevant need found is a precise definition of the toughness requirements for weldments produced with the HY-steels being examined. Should the specification exceed the actual need, an unnecessary burden has been imposed on fabricators. If the requirements can be eased slightly, fabrication costs could be reduced substantially and, with possible procedural changes, the resultant structures would be more reliable. Furthermore, if coarser heat-affected zones or slightly undermatched weld metals are shown to be acceptable, less expensive and more favorable welding procedures could be adopted.

In the final analysis, it will be necessary to determine the effects on overall cost effectiveness and structural integrity of weldments made with low-hydrogen electrodes and electrode alloys having mechanical properties that match those of the steels. Those in common use today overmatch the strength of the steels, contain "normal" amounts of hydrogen and, therefore, require preheat to prevent the development of hydrogen-induced cracks. The evidence on hand appears to confirm the logic behind eliminating preheat with the use of very low-hydrogen electrodes in combination with weld metal strengths that match or even undermatch those of the plate. This approach would reduce the added costs and complications associated with preheating weldments, however, without impairing the integrity of the completed structure.

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CHAPTER 8

STRENGTH OF WELDMENTS IN HIGH-STRENGTH STEELS

In the design and fabrication of welded structures, efforts are always made to ensure that various portions of a weldment have adequate strength and fracture toughness. In welding ordinary low-carbon steel, it is not difficult to obtain weld metals that match the base metals in both strength and fracture toughness (Masubuchi et al. 1966). In welding steels with higher yield strengths, however, it becomes increasingly difficult to obtain weld metals that give this match. This is particularly true of quenched and tempered steels, such as HY-80, HY-100, and HY-130, whose high yield strengths and excellent fracture toughness are obtained through heat treatment (Pellini, 1976).

There are basically two approaches to solving this problem. One is to accept weld metals that are undermatched to a limited extent in fracture toughness, but making certain that they are overmatched in strength to prevent strain concentrations and subsequent fractures in the weld zone when the weldment is subjected to tensile loading. The other approach is to accept weld metals that slightly undermatch the base metal in strength but have adequate fracture toughness.

The first approach has been used by the U.S. Navy in the construction of submarine hulls with HY-80 and HY-100 steels (Pellini 1977; Heller 1967). The adequacy of this approach has been proven by explosion bulge tests (Masubuchi 1966; Pellini 1976).

The second approach appears to have been incorporated in some recent Japanese standards following extensive research. Lower strength requirements may permit reductions in preheating temperatures and would result in easier welding of highly restrained heavy sections. In addition, with weld metals of lower yield strength, local plastic deformations can reduce stress concentrations in hard spots. Undermatched electrodes are often used for some layers (root and/or finishing) of multipass welding to prevent weld cracking. Investigators in different countries also have studied the properties of the weld metal itself and of the heat affected zone as each may affect the mechanical behavior of weldments in different materials (Bakshi and Shron 1962a, 1962b; Soete and Denys 1975).

Figures 8-1(a) and (b) show schematically how butt welds behave when they are subjected to tensile loading applied in two directions--perpendicular to the axis of welding and parallel to the axis of welding (Soete and Denys 1975). When tensile loading is applied to a transverse butt weld (Figure 8-1a), a weldment is under constant stress acting along the entire cross section. In such a case, when the applied stresses exceed the yield strength of the zone with the lowest yield strength, strain concentrations start to occur in that zone and result in fractures along the zone. However, the extent of the effects of the zone of lowest yield strength depends on the degree of reduction in strength and the width of the zone. When the tensile loading is applied to a longitudinal butt weld (Figure 8-1b), the weldment is under constant strain, and the effect of the zone of low-yield strength is very little as long as the zone has enough ductility and the width of the weldment, B, is reasonably larger than that of the weld metal, W. The effects of regions of lower strengths on the mechanical behaviors of weldments depend on several factors including: (1) types of loading (static tensile and compressive, repeated, etc.), (2) loading directions, (3) types of joint (butt weld of plates, fillet welds, etc), (4) sizes and geometries of weldments, and (5) the degree of reduction in strength and the width of the lower strength zone.

To determine the mechanical behavior of undermatched welded joints and to find reasonable strength levels of the filler metal from the standpoint of both workmanship and joint performance, the S. J. ("soft joint") Committee of the Japan Welding Engineering Society has undertaken extensive research during the past decade (Japan Welding Engineering Society, 1975). The fundamental studies (Satoh and Nagai 1969; Satoh and Toyoda 1970a, 1970b, 1971) were followed by detailed performance analysis of static tensile strength, fatigue strength, and brittle fracture strength of undermatched butt and fillet welds. The following discussions present, in summary, some of the results of these systematic Japanese studies. More details are given in Appendix C.

Tensile Strength of Undermatched Butt Welds

Fundamental Studies

The static tensile properties of welded plates--including soft interlayers and loaded either across or parallel to the weld line--were evaluated in extensive research completed by Satoh and Toyoda (1970a, 1970b, 1971, 1975) in the late 1960s and early 1970s at Osaka University. Some of this work is summarized in the following sections.

Round bar specimens of a medium carbon steel including a flash welded soft interlayer of low carbon steel were initially tested (Satoh and Toyoda 1970a). Results of the tests suggest that the strength of the joint approaches that of the base metal when the ratio of the thickness of the interlayer to the diameter of the bar is significantly small.

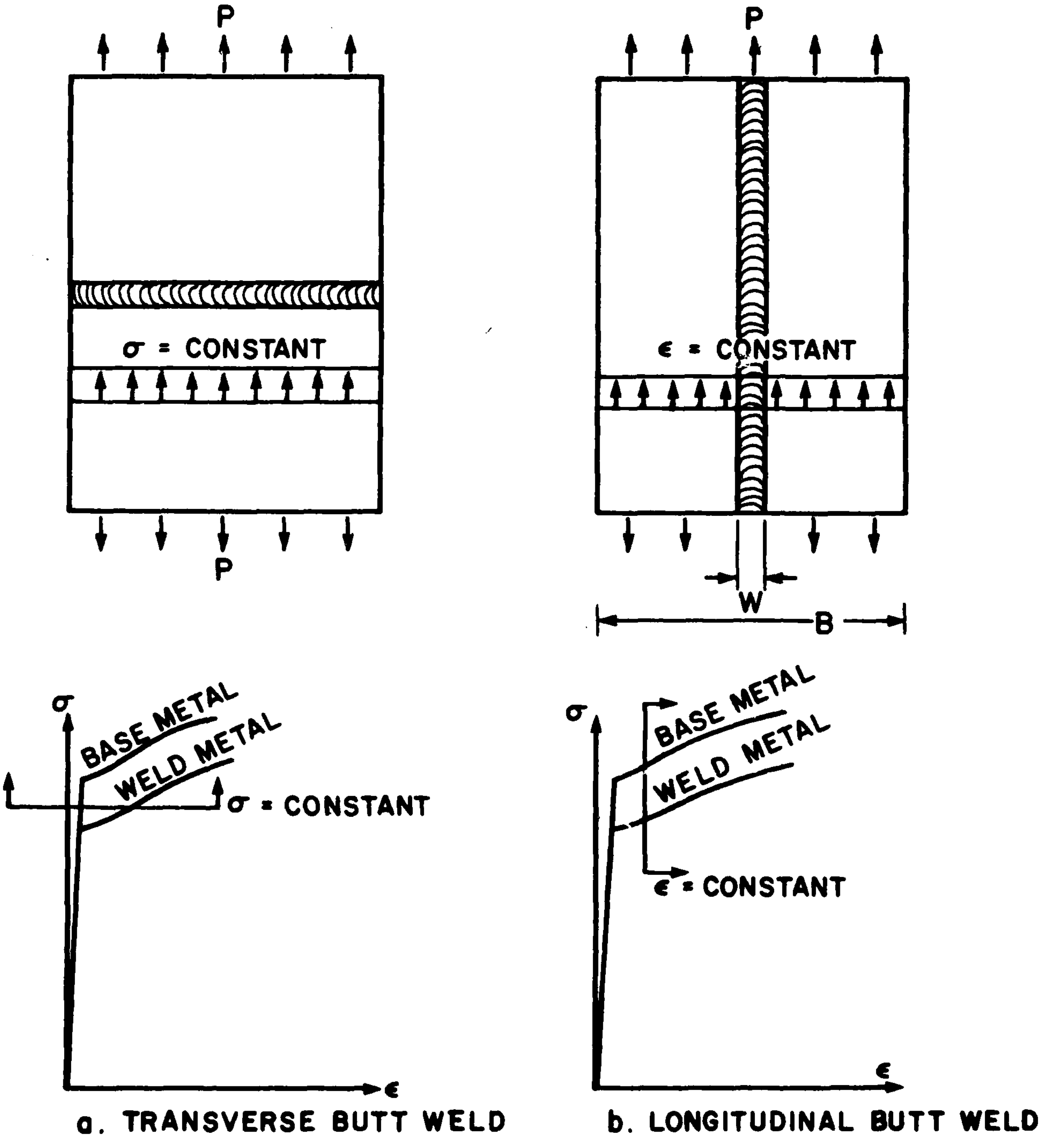


Figure 8.1 (a and b) Butt welds subjected to tensile loading.

Tests on flat bar specimens, loaded across the weld line (Satoh and Toyoda 1970b), indicate that the yield stress and the ultimate tensile strength of the specimens depend on both the relative thickness X_t (the ratio of the soft interlayer thickness to the plate thickness) and the plate width to thickness ratio (W/t). Specifically (Figure 8-2), the joint strength increases as the X_t decreases and reaches the strength of the base metal when X_t is small. Figure 8-2 also suggests that under constant X_t the ultimate tensile strength increases to a certain defined value that depends on X_t as the width to thickness ratio increases.

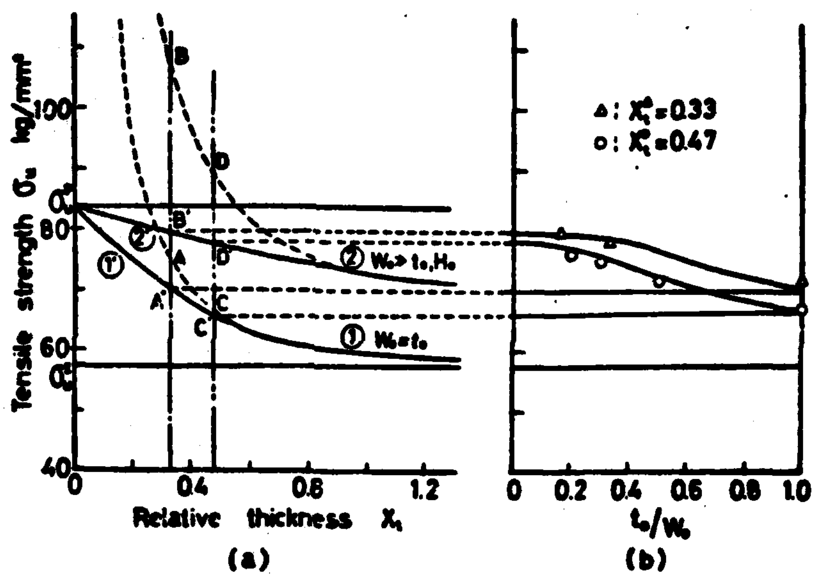


Figure 8-2 Effect of plate width on ultimate tensile strength of welded plates (Series S,T).

In experiments that simulate more applications experiments with loading parallel to the weld line (Satoh and Toyoda 1971), results suggest that the strength and ductility of the joint depend on the value of the ratio of the width of the hard zone to that of the soft zone. They become almost equal to those of the hard metal when the ratio is larger than 10 (see Appendix C for additional details).

Analytical and Numerical Investigation of the Strength of an Undermatched Joint

The experimental results discussed show good agreement with analytical predictions. Loading across a weld line (Figure 8-3) in cases of both infinite and finite width were investigated theoretically. When the joint width W_o is much larger than the thickness t_o and H_o, deformation in the width direction will occur in planes perpendicular to the x axis. Therefore, a plane-strain state can be assumed ($\varepsilon_x = 0$). For the analysis, it was further assumed that: (a) the welded joint consists of only two kinds of metals, base metal and a soft interlayer, each of which is homogeneous and isotropic and (b) the base metal behaves like a rigid body at any stage of loading and the contraction of the base metal under tension in the direction of thickness is negligibly small; (c) at a certain state of loading, the soft interlayer (the shaded portion in Figure 8-3) forms in such a way that there is no permanent change in volume; (d) the relation of equivalent stress $\bar{\sigma}$ to equivalent strain $\bar{\varepsilon}$ is of the form:

$$\bar{\sigma} = k\,(\bar{\varepsilon})^n$$

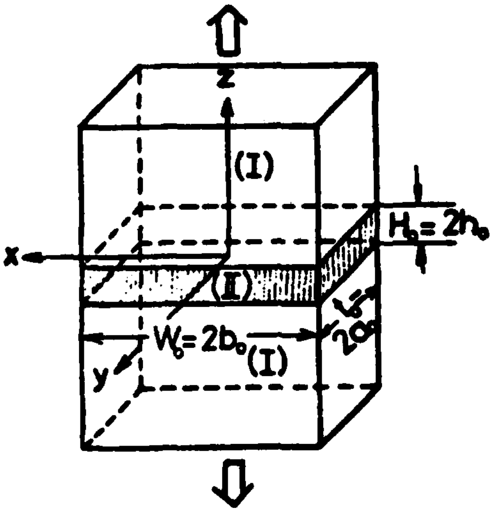


Figure 8-3 Welded plates including a soft interlayer $W_o \geq t_o$.

Further analysis of the multiaxial stress state in the neck region was performed in a way analogous to that employed by Davidenkov (1946). The results of Satoh and Toyoda (1970b) indicate dependence of the ultimate tensile strength of the joint on relative thickness, X_t or H_o/t_o, and are shown together with experimental data in figures of Appendix C.

The analysis for joints of finite width was performed under similar assumptions for specimens of both circular and square cross section. The results indicate that the ultimate tensile strength will depend on both relative thickness X_t and the ratio of the plate thickness to width t_o/W_o, as was already shown by the experimental data (Figure 8-2) in the previous section (Satoh and Toyoda 1970b). The analysis for loading parallel to the weld line was based on the further assumption that the strain distributions across a specimen were uniform at each load level (Satoh and Toyoda 1971). These results are also shown together with experimental data in the figures in Appendix C.

Although experimental results seem to verify the analytical predictions, an attempt was made to confirm the assumptions and results of the theoretical analysis with the use of the finite element program (ADINA) at MIT (Agapakis 1980). Both two-dimensional plane strain and axisymmetric analysis was performed that corresponded to the wide plate and round bar specimens tested by others. An elastic-plastic material model was used that assumes the existence of linear strain hardening and the Von Mises yield condition. To simulate a tensile test, a prescribed loading formulation was selected instead of a prescribed displacement and, to induce necking, a geometric imperfection was incorporated in the finite element mesh.

Various test cases were examined for a material similar to that examined by the Japanese investigators. Results (Figure 8-4) suggest that the applied load for fracture reaches that for the fracture of pure base metal at sufficiently small values of the relative thickness. Also, some of the assumptions of the theoretical analysis were verified; those that involve the deformation of the joint. The advantage of the finite element approach is that it can be used for the analysis of very complex joint geometries where simple theoretical analysis is impossible to perform.

Performance Study

The fundamental studies cited above indicated that for the idealized joints examined, the ultimate tensile strength may be as high as that of the base metal if the average width of the metal is sufficiently small.

To assess the applicability of undermatching joints in actual structures, the S. J. Committee of the Japan Welding Engineering Society carried out a performance study (Satoh and Toyoda 1971b, 1972; Satoh et al. 1979) presented in summary in Appendix C. The results indicated that for butt welded plate

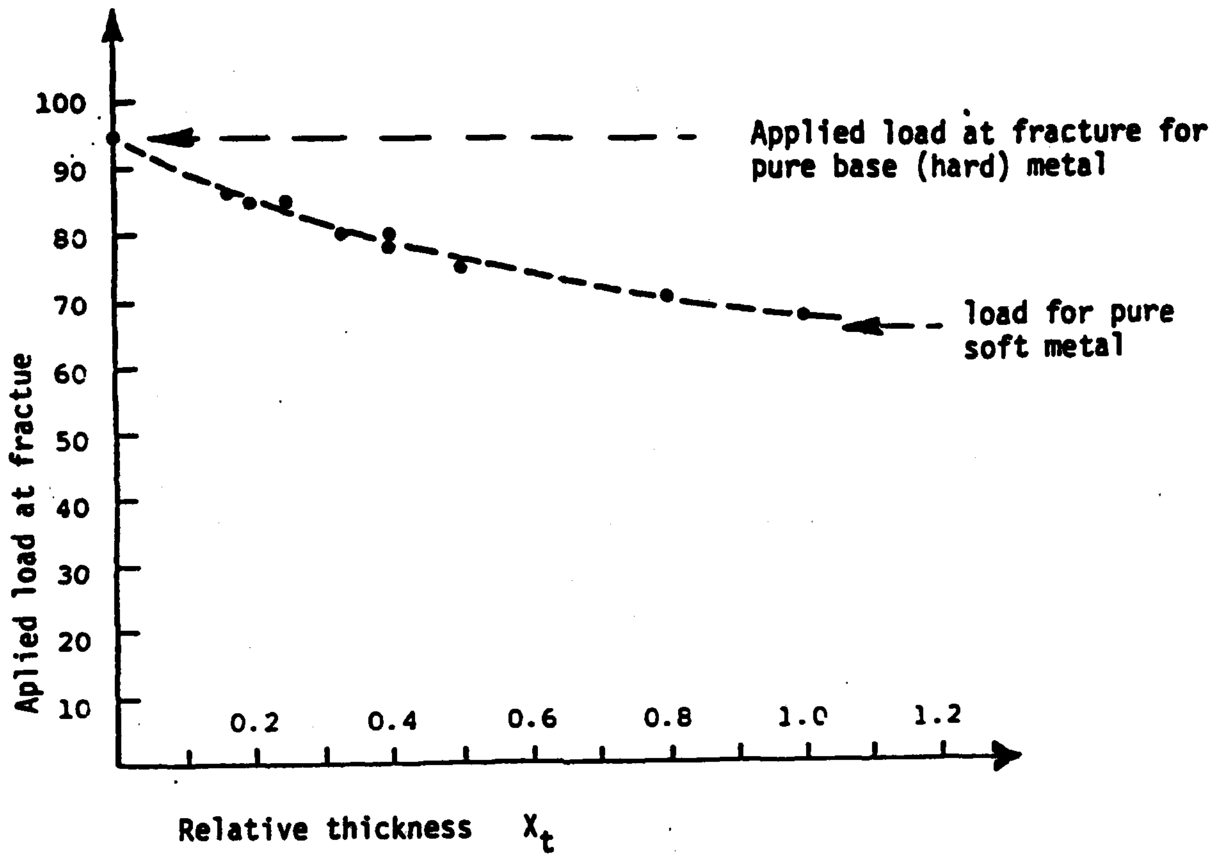


Figure 8-4 Applied load at fracture versus relative thickness (plane-strain case).

specimens, with an average relative thickness $(X_t)_{av}$ of 0.2 to 0.3, the strength of the joint reaches the strength of the base plate σ_u^B at nearly 90 percent of that of the weld metal σ_u^W. Thus, for a σ_u^W/σ_u^B ratio larger than 0.9, the undermatched weld joint behaves almost like the base metal in terms of both strength and ductility.

Tensile Strength of Undermatched Fillet Welds

The S. J. Committee of the Japan Welding Engineering Society also investigated the applicability of undermatching in fillet welds. Experiments were performed with various specimens of high-strength steel (U.T.S. of 84.1 kg/mm^2) welded with undermatched electrodes (U.T.S. of 40 kg/mm^2 to 80 kg/mm^2). The effects of various fillet geometries and the applicability of undermatching in repair welding were investigated. Some of the results are summarized in Appendix C. A detailed presentation as well as results for shear strength of undermatched fillets is given by the Japan Welding Engineering Society (1975). The pertinent conclusions from this work are given in each of the review sections below and in the summary at the end of this chapter.

Fatigue Strength of Undermatched Welded Joints

Butt Welds

Gelman and Kudrayavtzev (1964) showed experimentally that the fatigue strength of bars with a soft interlayer increased when the thickness of the interlayer decreases. In the late 1960s, Satoh and Nagai (1969) investigated the fatigue strength of bars having either hard or soft interlayers. Some discussion of these experiments is given in Appendix C. In summary, it was found that the hard interlayer had no effect on fatigue strength of the specimen, whereas for the soft interlayer, the fatigue limit decreases drastically as the thickness of the interlayer increases. The performance of undermatched welded joints was evaluated by the S. J. Committee with a series of fatigue tests on various types of high-strength steel specimens welded with under- and overmatching electrodes. The results indicate that undermatching has a small effect on the fatigue strength for loading transverse to the weld line. For loading parallel to the weld line, however, the fatigue life of the overmatched joint is somewhat longer.

Fillet Welds

The S. J. Committee also investigated the fatigue strength of undermatched fillet welds and noted the effect of the shape of the fillet (Japan Welding Engineering Society, 1975). The conclusions were similar to those drawn from the work of Ikeda and coworkers (1975). The authors developed a new electrode that improved the toe geometry attainable with the conventional one. Fatigue tests indicated that the crack-initiation life almost doubled in the specimen welded by the new electrode, whereas the grinding of the tow surface hardly influenced the fatigue limit.

Residual Stresses in Weldments in High-Strength Steels

It is an established fact that welding causes residual stresses. The magnitude and distribution of these stresses depend on factors such as the geometry of the weldment, the properties of the weld and base metals, and welding processes and procedures. A clear understanding of residual stresses is not yet available but progress in this area has been made through analytical and theoretical studies. Presented here are summaries of (1) recent advances in the analysis of residual stresses in weldments and (2) experimental and analytical information on residual stresses in weldments in high-strength steels, especially quenched and tempered steels.

Analytical Models for Residual Stresses

The role of analytical models in understanding residual stresses is important for both economic and technical reasons. Economically, it becomes expensive to depend only on experimental studies to evaluate residual stresses caused by welding. After a degree of confidence is obtained in the predictive

capability of the analytical model, it can be used to complement the experimental effort by indicating which are the most beneficial experiments to be conducted. Technically, the analytical model offers a representation of the trends associated with changes in a single parameter being studied. This information is often not easily obtained experimentally. The model also provides a way to extrapolate from a given set of results to new conditions.

Since the 1960s, a number of investigators have conducted computer-aided analyses of residual stresses in weldments. Masubuchi and coworkers (1968) and Vaidyanathan and coworkers (1973) present one-dimensional models for residual stresses caused by welding. Two-dimensional finite element models for predicting weld-induced residual stresses have been presented, for example, by Friedman (1975), deYoung and Chin (1977), Ueda and Yamakawa (1971), Nomoto (1971), Muraki and coworkers (1975), and Rybicki and coworkers (1978). These models focused on a small number of weld passes. Rybicki and Stonesifer (1979) have developed a finite element model for multipass welds that handles up to 30 passes and also a model for a weld repair with more than 900 weld passes. Good agreement was found between measured residual stresses and those computed by the model. Further studies show the feasibility of using the analytical model to evaluate methods of controlling residual stress by altering the welding process or by postweld heat treatment (Rybicki and McGuire 1980). These models have been used successfully for steels and aluminums of low yield strength. Much less analysis has been done for high yield-strength materials, and this topic is discussed in the following section.

Longitudinal Residual Stresses

It has been well established that high-tensile residual stresses usually exist parallel to the weld line in regions near the weld. In weldments made with low-carbon steels with yield strengths of 35 to 40 ksi, the peak residual stress usually approaches the yield stress. Curve 0 of Figure 8-5 shows schematically a typical distribution of longitudinal stresses, or stresses parallel to the weld line, in a butt weld in low-carbon steel. Residual stresses are tensile in regions near the weld, but become compressive in regions away from the weld. The information on the magnitude and distribution of residual stresses in weldments in high-strength quenched and tempered steels, however, is still limited and inconclusive. It has been established that residual stresses are generally tensile near the weld and then become compressive in regions away from the weld. The critical questions are:

(1) Are peak residual stresses as high as the yield stresses of the weld metal and the base metal?

(2) How wide are the areas where high-tensile residual stresses exist?

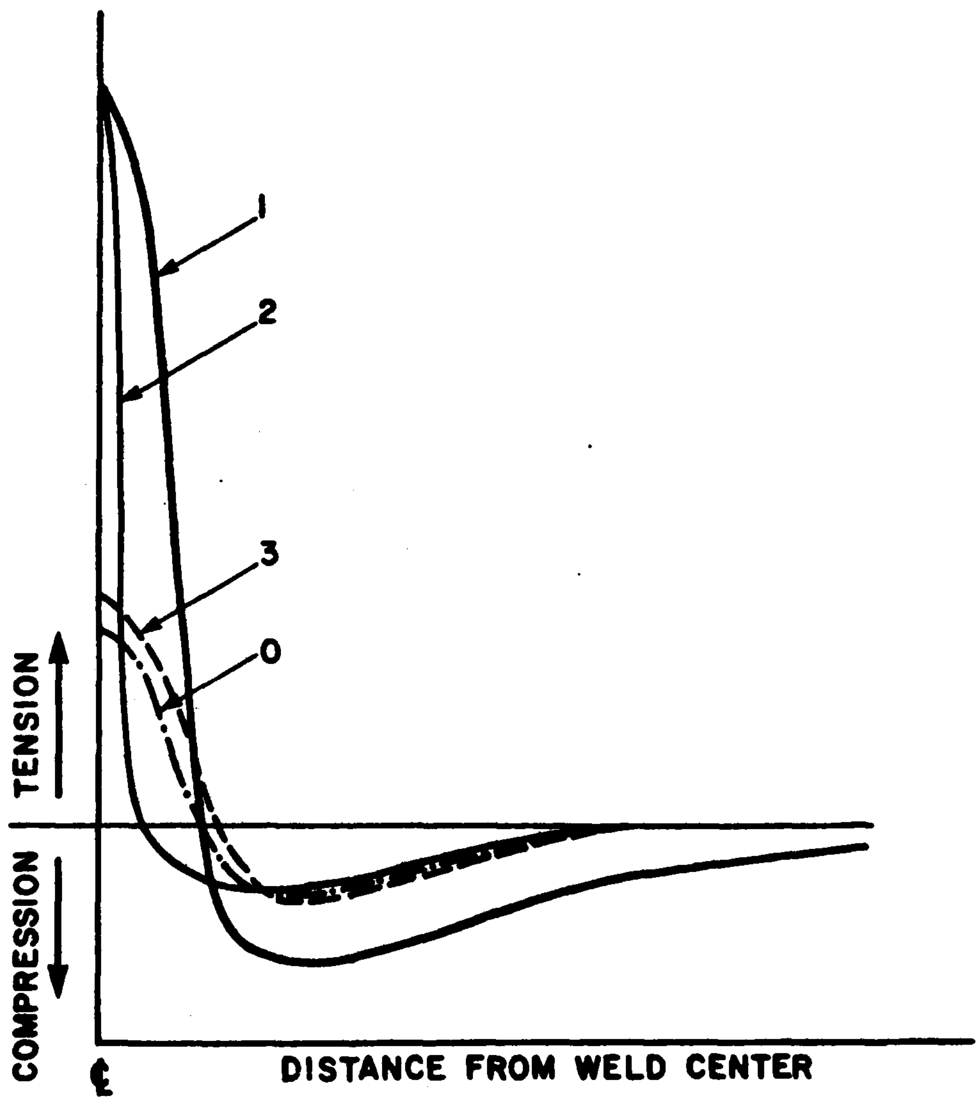


Figure 8-5 Possible distributions of longitudinal residual stresses in a butt weld in high-strength steel.

Curves 1, 2, and 3 of Figure 8-5 show three possible distributions of longitudinal residual stresses in a butt weld in high-strength steels. If it is assumed that the maximum residual stress is as high as the yield stress, the distribution would be given by Curve 1. If this were true, the residual stress and distortion would cause severe problems in the fabrication of welded structures using high-strength steels. In Curve 2, the high-tensile residual stresses are confined to small areas. In such a case, the distortion would be less, but cracking from high tensile residual stresses would be a problem. The stress distribution given by Curve 3 would cause few problems.

Masubuchi's (1980) recent book presents some experimental and analytical results on residual stresses in weldments in high-strength steels. No existing experimental and analytical data support Curve 1. Fabrication experience of welded high-strength steel structures also supports the possibility that Curve 1 can be ruled out. Experimental data obtained with strain gages and sectioning techniques tend to support Curve 3 (Masubuchi 1977; Akita and Yada 1965; Yurioka 1972). Some experimental data indicate that residual stresses could be quite low in some regions of the heat-affected zone in weldments in certain high-strength steels (Adams and Corrigan, 1966; Wohlfart 1976). A possible explanation, suggested by several investigators, is the effect of solid-state transformation during cooling. However, recent MIT experimental data in heavy section HY-130 steel weldments show that the peak residual stress can be as high as the yield stress (Papazoglou and Masubuchi 1980).

On the other hand, analytical data that neglect the effect of phase transformation tend to support Curve 2 (Masubuchi 1980). Efforts have been made in the MIT study to include the effect of solid-state transformation in the analysis of thermal stresses and residual stresses in high-strength steel weldments (Papazoglou, 1981). Although the analysis indicates that residual stresses in some regions near the weld can be reduced considerably by the expansion caused by transformation, it is still premature to make firm statements on residual stress distributions in weldments in quenched and tempered steels.

On the basis of the experimental and analytical information obtained thus far, it appears that residual stress distributions in actual structures are mixtures of Curves 2 and 3. Peak residual stresses in some areas of the weld metal and possibly in the heat-affected zone in some cases could be as high as the yield stress. However, average values of residual stresses could be close to Curve 3. If x-ray diffraction techniques are used for measuring residual stresses in the weld metal and heat-affected zone, it is likely that widely scattered results, ranging from the yield stress to much lower values, will be observed.

Regarding the effect of overmatching or undermatching, it is believed that welds made with higher-strength weld metals have higher residual stresses than welds made with lower-strength weld metals.

Transverse Residual Stresses

High tensile residual stresses do not normally exist perpendicular to the weld line unless welds are highly restrained (Masubuchi 1980). However, high tensile transverse residual stresses can occur in various cases, e.g.:

(a) When only a portion of a restrained butt joint is welded, very high tensile residual stresses can occur in the weld metal. In fact, the stresses may well exceed the yield stress, and the weld may fracture, resulting in weld-metal cracking.

(b) In welding a heavy plate using the multipass procedure, high tensile residual stresses may be caused in some portions of the joint. For example, welding of the last layers on one surface of a plate can cause high-tensile residual stresses in regions near the other surface of a plate.

It is believed that welds made with lower-strength weld metal have lower residual stresses and thus are less prone to cracking than welds made with higher-strength weld metal, although there have been no published data to substantiate this comment. Treatments such as preheating and postheating will have some effects on residual stresses and crack sensitivities of overmatched or undermatched welds. Again, there are no published data. Systematic analytical and experimental investigations of these subjects are needed.

Fracture Resistance

Effects of Residual Stress on Fracture

The extent to which residual stresses affect the fracture strength of a weldment depends greatly on the brittleness of the material. When the material is brittle and fractures occur under low applied stress, residual tensile stresses can significantly reduce the fracture strength of weldments. Many investigators have studied the effects of residual stresses on brittle fracture of welded structures (Masubuchi 1977; Wells 1956; Kihara and Masubuchi 1959; Hall et al. 1967). Residual stresses also play important roles in stress corrosion cracking and hydrogen-induced delayed cracking of weldments that occur even without external loading (Masubuchi and Martin 1966).

It is well known that the effect of residual stress is practically nil when fracture occurs after general yielding, either in a brittle or ductile manner.

Fracture Strength of Undermatched Welded Joints

Various investigations examined the fracture strength of undermatched welded joints in high-strength steels (Satoh et al. 1972, 1977; Satoh and Toyoda 1973). Some of these results are presented in Appendix C. It was

concluded from these results that, generally, higher fracture toughness or lower transition temperature should be required for the undermatched weld metal than for the overmatched one from the standpoint of brittle fracture initiation.

If T_{vs} is the fracture transition temperature obtained from V-notch Charpy test and $\Delta_v T_s$ is the difference of the fracture transition temperature between overmatching and undermatching filler metals required to obtain the same fracture initation temperatures, T_i, analysis has shown that

$$\Delta T_{vs} = 80 \ln(S_r)_y[1-65(1/T_i - 1/273)]$$

where $(S_r)_y$ is the ratio of the yield stress of the undermatching to that of the overmatching filler metals. For example, if $(S_r)_y$ is 0.8 and T_i is 50° to -150°C, then the required T_{vs} in the undermatching is about 15° to 20°C less than that for the overmatching one.

Structural Applications

From the early fundamental study and the joint performance tests carried out by several Japanese investigators, it became clear that the undermatching electrode can be applied in practical welding of heavy plates of high-strength steels. The experimental and practical data are summarized below.

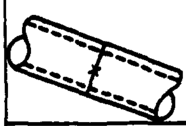
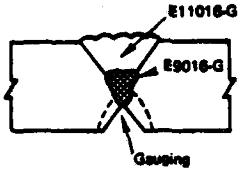
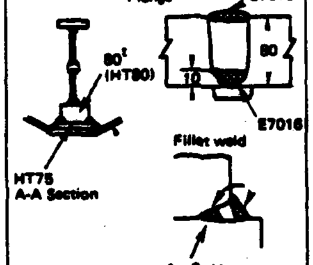
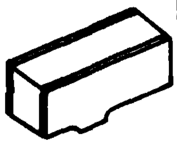
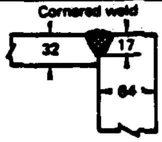
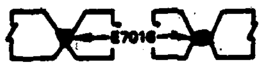
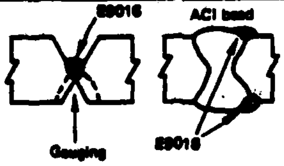
An early example of experimental work was a burst test of welded pipes of HT-80 steel plate. (HT should not be confused with HY. The HT steels referred to in this report is the Japanese designation for high-tensile strength steels and have no direct relation to the HY steels used in the United States.) Welds were made with an electrode having an ultimate tensile strength (UTS) of 77 kg/mm^2. However, during the burst test, fracture started at an internal pressure of 235 kg/mm^2, corresponding to a circumferential stress of 90 kg/mm^2 (just above the minimum ultimate tensile strength of the base metal, 89 kg/mm^2). It is apparent, therefore, that the undermatched filler metal had no harmful influence. Examples of undermatched welded joints in actual structures in Japan are given in Table 8-1.

An extensive investigation of the potential applicability of undermatching welding of HT-80 heavy plates for penstock was also carried out in Japan (Satoh et al. 1979, 1978a, 1978b). The most important conclusions from both the experiments and the service experience are the following:

- The undermatched E9016G electrode effectively lowers the preheating temperature required for preventing root crack caused by the first pass welding of 50 mm thick HT-80 steel plates by around 25°C compared with the overmatched E11016G electrode.

- Weld metal cracking caused by continuous multipass welding is not as harmful to joint restraint severity, and the preheating temperature required for preventing weld-metal cracking can be lowered more than 50°C or 75°C by using the undermatched E9016G electrode.

TABLE 8-1 Some Examples of the Undermatching Welded Joints in Actual Welded Structures

Structure	Kind of Welded Joint	Detail of Welded Joints	Used Material (thickness)	Welding Method and Electrode	Remarks
Penstock	Circumferential butt welding 		HT-80 (32mm)	Shieldmetal-arc welding Weld metal AWSE9016-G AWSE11016-G	1973-1975 Pumped storage power station
Penstock	Branch  Arch girder		HT-80 HT-75 (32-80mm)	Shieldmetal-arc welding Weld metal AWSE7016 AWSE11016-G	1972 Power station
Bridge	Box girder 	Cornered weld  Fillet weld of rib 	HT-80 HT-70 (32-64mm)	Gas metal arc welding or submerged arc welding Weld metal for HT80	1974 Bridge in Osaka
	Tower	Groove weld 1st pass / after gauging 1st pass  Fillet weld		Shieldmetal-arc welding Weld metal AWSE7016 AWSE11016-G for HT80	
Pressure Vessel	Butt welding		HT-80 (25mm)	Shieldmetal-arc welding Weld metal AWSE9016-G AWSE11016-G	

- Appreciable differences in tensile strength and uniform elongation between the overmatched and undermatched weld joints are not observed in welded wide plate tension specimens, with or without notch, tested at a temperature slightly lower than the minimum service temperature experienced in the penstock.

- To satisfy a safety requirement for circumferential connections of penstock of HT-80 steel plates, whose tensile strength ranges between 80 kg/mm^2 (114 ksi) and 90 kg/mm^2 (128 ksi), the tensile strength of weld metal deposited by undermatched electrode should not be less than 65 kg/mm^2 (92 ksi).

Summary

When welding high yield strength steels, it becomes very difficult to obtain weld metals that match the base metals in both strength and fracture toughness. Contrary to the U.S. specifications, Japanese industrial standards accept weld metals that slightly undermatched the base metal in strength but have adequate fracture toughness. Extensive studies have been performed by various investigators to evaluate the performance of such undermatched weld joints. Some of the analytical and experimental results presented in this chapter and Appendix C, are summarized below:

<u>Tensile Strength of Undermatched Welded Joints</u>: Initial experimental studies with butt welded specimens established that the strength of an undermatched joint loaded in a direction transversed to the weld line approaches the strength of the base metal when the ratio of the thickness of the soft interlayer to the thickness of the plate is significantly small. These trends were also observed with larger plate specimens and have been verified by various analytical studies.

<u>Fatigue Strength of Undermatched Welded Joints</u>: The fatique strength of welds is governed either by the shape of the weld or the strength of the weld metal, depending on how the fatigue crack is initiated. When the shape of the weld is the governing factor, the strength of the weld metal has little effect on the fatigue strength of the joint. When the strength of the weld metal is the governing factor, the fatigue strength of the joint increases with the strength of the weld metal.

<u>Residual Stresses in Weldments in High-Strength Steels</u>: It has been established that in weldments made with low-carbon steel, residual stresses as high as the yield stress usually exist parallel to the weld line in regions near the weld. However, information on the magnitude and distribution of residual stresses in weldments in high-strength quenched and tempered steels is still limited and inconclusive. Some experimental data indicate that peak longitudinal stresses can be as high as the yield stress of the material. Transverse residual stresses, on the other hand, are considerably lower than the yield stress unless the weld is highly restrained.

Systematic analytical and experimental investigations of the residual stresses in weldments of quenched and tempered steels are still needed in order to study the effect of the yield strength of the weld metal and preheating.

Fracture Resistance: From the standpoint of fracture prevention, higher fracture toughness or lower fracture transition temperature should be required for undermatched weld metal than for overmatched weld metal.

Applications: The results of the fundamental experimental and analytical studies in Japan encouraged the use of the lower strength electrode materials in actual structural applications. In most cases the undermatched electrodes required lower preheating temperatures and resulted in no significant loss of strength or ductility.

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CHAPTER 9

DISCUSSION

Traditionally, codes and fabrication documents have required that a weldment possess properties as good as or better properties than those of the base plate in structural fabrications. In almost all cases, the weldment must exhibit matching or superior strength and "adequate" toughness and resistance to stress corrosion cracking. Weldments with inferior strength have been acceptable only in a few limited cases--repair welds in HY-80 (made with covered electrodes of the type E-9018) and in a recent fabrication of girth welds of penstocks where the welds are never exposed to maximum pressure stresses. The use of weld metals of "less than matching" strength has never been put to the test. Certain advantages could be achieved if weld metals with lower yield strengths than the base metal were able to perform satisfactorily in high-strength steel weldment systems.

The economic advantages in fabrication include lower sensitivity to hydrogen-induced cracking, reduction of preheat requirements for HY-systems, and the use of higher deposition rate processes. A survey of fabrication economies for HY weldment systems made certain conclusions clear. Fabrication costs can be reduced by (1) removing preheat requirements for HY-steel systems and (2) allowing higher deposition rate processes to be used.

Fabrication problems increase as the required strength of the weld metal increases because more careful control of preheat, interpass temperature, heat input, and hydrogen is required. Preheating and control of interpass temperature are expensive and are required in all cases where HY steels are welded. In some, less critical situations, it is difficult to justify this extra cost. Further, any added confidence in the ability to minimize weld defects (e.g., hydrogen-induced cracking) would naturally lead to less stringent inspection requirements.

The major objection to relaxing preheat and interpass temperature requirements for HY-steel welding is the threat of hydrogen-induced cracking--a phenomenon that introduces to the weldment an undesirable defect in the heat-affected zone or the weld metal. Time-dependent hydrogen-induced cracking that occurs between -100°C and 200°C is promoted by a combination of the following factors:

1. A sensitive, usually martensitic, microstructure--whose hardness generally exceeds R_c30.

2. The presence of tensile stresses--usually the residual stresses in the weldment are adequate.

3. The presence of hydrogen which is common in the welding of steels.

4. A holding temperature between -100°C and 200°C.

The absence of any one of the above factors can eliminate the possibility of hydrogen induced cracking.

The sensitivity of HY-steel systems to hydrogen-induced cracking increases as strength increases. At the HY-80 level, cracking is limited mainly to the heat-affected zone of the base metal, but in the stronger HY steels the cracking problem transfers to the weld metal. Even if cracking is not encountered in the weld metal, the hydrogen problem shows up in a standard tensile test as a reduction in ductility of the weld metal. Whereas an elongation of 14 percent has been established for HY-130 weld metal, the presence of hydrogen can result in lower values. HY-80 weld metal, however, does not exhibit this same sensitivity to hydrogen.

Control of the hydrogen damage situation, either cracking or low ductility, must be achieved by manipulating the factors that cause the problem. Minimizing the hydrogen content of the weldment is known to reduce the likelihood of damage. Few problems exist when welding is done with the gas metal arc welding (GMAW) process, where the chance of hydrogen absorption in the weld is reduced. The shielded metal arc welding (SMAW) process has traditionally had problems because the coating of the electrodes has contained hydrogen, in the form of either absorbed moisture or the water of crystallization of the components. The submerged arc welding (SAW) process suffers from similar problems. Recently, some manufacturers of SMAW consumables have developed "low moisture content" electrodes that absorb moisture at minimal rates. The hydrogen content of welds made with these electrodes is very low. Nevertheless, it is suspected that the amount of hydrogen required to erode the properties of the weldment in high-strength steel is very low--1 ppm has been shown to be sufficient. For example, HIC in HY-80 has been found with weld-metal hydrogen content in the order of 3-5 ppm.

The relation between the hydrogen potential of the welding process and the amount of hydrogen in the weldment is not well established because of the different ways in which hydrogen can be introduced. Water adsorbed on an electrode coating may be driven off before entering the arc region of the electrode, whereas water of crystallization is stably contained in the coating and, therefore, probably is more effective in introducing hydrogen to the weld. However, the "hydrogen potential" of welding processes is becoming more carefully controlled.

Strength affects the weldability of HY steels in two ways. First, the strength of the weld metal is dictated by the microstructure of the weld metal. The microstructure is again determined mainly by the composition of the steel and the thermal history; usually by the cooling rate from the

austenitic region. To achieve a certain strength, the cooling rate of the weld metal must be controlled to produce either a bainitic structure in the HY-80 or HY-100 system or a martensitic structure in the HY-130 system. The predetermined cooling rate then determines the welding parameters such as preheat, interpass temperature, and heat input. The higher the strength required, the more critically controlled must be the welding variables.

Important to the problem of hydrogen-induced cracking is the belief that strength influences residual stress in weldments. The higher the yield strength, the higher is the potential residual stress and hence the sensitivity to hydrogen. This point is not well established experimentally, although it is believed that some localized stresses can be as high as yield strengths, even for HY-130 steels. No effect of preheating on the magnitude or distribution of residual stress has been documented in the literature. A reduction of yield-strength requirements would tend to reduce the likelihood of hydrogen-induced cracking, both from the point of view of residual stress and microstructural sensitivity.

Thus, present knowledge would favor weld metal of reduced yield strength. The major remaining question relates to the performance of weldments made with weld metal of matching or undermatching strength. Again, we must confess that little experience is available to support a recommendation that low-strength weld metals will be adequate for service conditions. On the other hand, if one assumes that a structure is designed on the basis of the yield or ultimate strength of the base metal, there is no evidence to suggest that any advantage can be gained from a weldment whose weld-metal strength overmatches that of the base metal. Even if one assumes that the weld is the most likely location for defects to occur, there is little evidence that a 10 percent overmatch of weld metal strength will reduce the possibility of failure in the weld. This is particularly the actual case because in a given steel weldment system, the toughness of the weld metal tends to decrease as its strength increases. In fact, a review of HY-steel specifications does not reveal a logical toughness requirement for weldments, and criteria of acceptance vary from one welding process to another.

Very limited data indicate that adequate strength can be achieved in butt weldments with a 10 percent undermatched weld metal, provided that the joint is designed to take advantage of triaxial constraint. Undermatched fillet welds present little problem because the size of fillet welds can be adjusted to compensate for the shear strength of the joint. Whether increased toughness should be required for undermatched weld metal is not clear and deserves more research. It can be argued whether an undermatched weld metal should be able to tolerate defects better when exposed to greater strain (or stresses of greater percentage of yield stress). From the point of view of fatigue, there is no evidence to suggest that an overmatched weldment will resist cyclic crack growth better than will a matched or undermatched weldment. The lack of evidence is reasonable because of the the predominant effect of geometry (stress raisers or weld ripples) on fatigue performance.

In brief, there is little evidence that one must overmatch a high-strength steel weldment system to achieve adequate performance or that slight undermatching will not provide adequate performance. Tradition indicates that one must overmatch--the matched or undermatched situation has not been adequately considered.

The technology of manufacturing consumables permits the strength of weld metals to be predicted for a given joint thickness and cooling rate. Therefore, consumables can be specially formulated to give reliable levels of strength--matching or controlled undermatching--particularly for the HY-80 and HY-100 weldment systems. For HY-130 weldment systems, a lower yield-strength requirement would facilitate the qualification of covered electrodes. It would also allow increased reliability of processes with higher deposition rates, such as submerged arc welding, which is very marginal based on currently required mechanical properties.

This review brings the committee to recommend most strongly that renewed consideration be given to matching or undermatching weld metal strength in high-strength steel weldment systems. The potential rewards are technical and economic and involve:

1. Increased structural reliability can be enhanced by decreased sensitivity to hydrogen-induced cracking due to

    - Lower residual stresses.
    - Less sensitive microstructures.
    - Lower "hydrogen potential" of improved consumables.

2. Cost effectiveness related to

    - Possible elimination of preheat requirements for weldments having low restraint.
    - Relaxed inspection requirements based on greater confidence that weldments will be free of defects.
    - Greater use of higher deposition rates at high strengths (HY-130).

The matching or undermatching weld metal system should be further investigated experimentally. It should then be tested on noncritical joints to gain experience and confidence before radically changing procedures (such as preheat requirements) for weldments in critical structural members.

APPENDIX A

RESUMES OF COMMITTEE MEMBERS

A short professional resume of the committee members is given below.

David G. Howden, Chairman of the Committee, received his B.SC (1959) and Ph.D. (1962) degrees in metallurgy from University of England. He was assistant professor and research metallurgist at Cent. Tech. Aeronaut. in Sao Jose dos Campos, Brazil (1963-65) and senior scientist of welding metallurgy in the Depatment of Energy, Mines & Resources, Canada (1965-67). He joined Battelle Columbus Laboratories as associate manager of welding and fabrication (1967-77), then joined the faculty of the Department of Welding Engineering at Ohio State University as associate professor in 1977. His professional expertise is in welding metallurgy, gas-metal reactions in arc welding, health and safety in welding, welding process development, and component failure analysis

Bruno L. Alia received a B.S. (1971) in mechanical engineering from The Cooper Union. He served on the Metallurgy Staff at the American Bureau of Shipping (1967-78) and is now chief surveyor and head of metallurgy department. His professional activity has been in mechanical and metallurgical engineering, production welding, nondestructive testing, welding process development for high-pressure components, and materials selection for marine structures.

James M. Cameron attended the Lincoln School of Welding (1940) and received a B.S. degree in electrical engineering from the Unviversity of Buffalo in 1955. He worked as project engineer at Westinghouse Electric Co. (1947-55) and as supervisor of welding research and production at ACF Industries (1955-63). He is currently manager of materials and welding engineering at the Electric Boat Co., General Dynamics Corp. His expertise is in welding process development for critical structures and materials engineering and weldability studies for forgings and castings.

John J. Koziol received a B.Met.E. from Rensselaer Polytechnic Institute (1959), and a M.S. in metallurgy (1968) and a M.B.A. (1971) from the University of Connecticut. He worked as research engineer at United Nuclear Corp. (1961-62) and as senior analytical engineer at Pratt & Whitney Co. (1962-64). He joined Combustion Engineering, Inc. in 1964 as development engineer and is now manager, systems materials, Nuclear Power Systems Division. He has worked in the areas of mechanical properties, fabrication, materials selection, and design review.

Alexander Lesnewich, Airco, Inc., received a B.S. degree (1948) and a Ph.D. degree (1952) in metallurgy from Rensselaer Polytechnic Institute. He served as senior engineer in the research laboratory of Air Reduction, Inc. (now AIRCO) (1952-57) and is currently director of research and development of Airco Welding Products Division. His professional contributions have been in the fields of inert gas arc welding, welding processes, and welding metallurgy.

Benjamin C. Howser, Newport News Shipbuilding & Drydock Co., attended the University of Miami majoring in government and business law. He has been with the Newport News Shipbuilding & Drydock Co. since 1961 and has moved through a number of supervisory positions to become manager of the welding engineering department (1974). He has been involved in all phases of welding including production welding, manpower planning, nuclear structure welding, and welding technology improvement.

Ernest F. Nippes, Professor at Rensselaer Polytechnic Institute, received a B.S. degree (1940) and a Ph.D. degree (1942) in metallurgy from RPI. He stayed on as a faculty member of the metallurgy department as instructor and became professor of metallurgical engineering in 1954. He has directed the departments efforts in welding research and has served as department chairman and as director of the office of research and sponsored programs. His professional expertise covers the broad scope of metallurgy and welding.

Edmund F. Rybicki received a B.S. degree in civil engineering (1963), a M.S. (1965) and a Ph.D. (1968) in mechanical engineering from Case Institute of Technology. He worked as engineering scientist in the Advanced Solid Mechanics Section, Battelle Columbus Laboratories and moved up to projects manager (1968-79). He joined the faculty of the University of Tulsa in 1979 as professor and chairman of the Department of Mechanical Engineering. His areas of specialization are mechanical properties of materials, pressure vessel weld repair assessment, fracture mechanics and residual stress analysis of weldments, and computational fracture mechanics.

Robert C. Shutt received a B.A. degree in physics from Kenyon College in 1950. He joined Lincoln Electric Co. as development engineer (1958) and is currently Vice President, Electrode Development. His professional activities encompass the areas of weld metal research and development, submerged arc fluxes and self-shielded flux cored electrodes, high-alloy steel filler metal evaluation, and electroslag and electroslags filler metal development.

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OHIO

Your reference ----
Our reference N6E5.2
Date 8 August 1980

Copy to: Mr R Cooper, DGS 123, Block B, Foxhill, Bath

Dear Professor Howden

I understand from my colleague John Bird (and also see from the Welding Research Council Progress Report) that you are chairing a study group into yield and toughness considerations for high strength steels in the HY80 to HY130 strength range.

This is a topic of considerable interest to us and is an area where we have several research projects underway. As regards weld metal to plate yield strength ratio, we are still using an overmatching philosophy, but this is coming under increasingly close scrutiny.

For a structure which sees only elastic stresses there is obviously little point in overmatching yield stress, indeed the effects may be deleterious from the residual stress viewpoint. The only possible justification for overmatching is to prevent concentration of strain in the weld under conditions of severe plastic deformation. However, the tougher the weld the less necessary it is to shield it from plastic deformation; and it seems likely there is a trade-off between strength and ductility or toughness which may not be optimised by the overmatching situation.

In a recent series of bulge explosion tests to assess the effect of undermatching we used the same 650 MN/m^2 consumable with HY80, HY100, and HY130 plate. Results were satisfactory in both the HY80 and HY100 plate, but disastrous in the HY130. In the HY100, which represented something like a 10% undermatch on yield, strain enhancement did occur in the weld. After 4 shots, when strain in the plate was 12%, strain in the weld was 22%. After 3 shots at the HY80 strength level (approximately 10% overmatch) strain in the weld was 15% compared to 13% in the plate. In neither case were any defects seen in the weld. Were pre-existing defects to be present in the weld, the enhancement of weld strain in the undermatching case would obviously cause fracture at earlier plate strain than in the overmatching case (assuming the same fracture toughness in the weld). This is, however, likely to be offset by the higher fracture toughness normally obtainable in lower yield strength welds.

Of course, if high toughness and ductility can be obtained at the same time as overmatching yield strength, there is no incentive to undermatch. The main problems seem to arise with manual metal arc welding. With automatic processes, high toughness and yield can usually be achieved simultaneously. Results with synergic pulsed arc MIG seem especially good. If this can be developed to a stage where it is a viable alternative to MMA for main constructional welds the undermatching/overmatching argument may become irrelevant.

Until then, however, we are extremely interested in the findings of your study group and should be in a position to make a more formal contribution sometime within the next twelve months.

I enclose a general paper on toughness considerations for high strength weld metals. We have, in addition, a considerable amount of fracture toughness data on HY80 strength level welds if this would be of interest to you.

Yours sincerely

John Sumpter

J D G Sumpter

JDGS/BG

APPENDIX C

EXCERPTS FROM THE LITERATURE OF THE JAPANESE STUDIES CONCERNED WITH WELD-METAL STRENGTH UNDERMATCHING

During the past decade, the Soft Joint (S.J.) Committee of the Japanese Welding Engineering Society made an extensive investigation of the mechanical behavior of undermatched weld joints. Chapter 8 briefly presented some of the results and more details on this effort are given here to explain more fully the findings.*

Tensile Strength of Undermatched Butt Welds

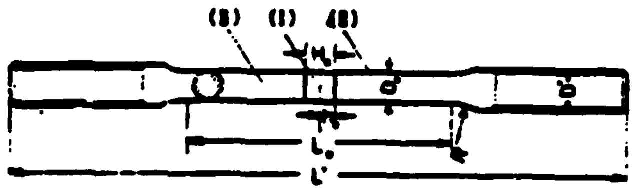
Fundamental Study

A series of tests were made to evaluate the static tensile properties of welded plates. The results are summarized below.

(A) Round bar specimens of a medium carbon steel (Japanese standards: S35C), including a soft interlayer of low-carbon steel (S10C), were initially tested (Satoh and Toyoda 1970a). The specimen shown in Figure C-1 was prepared by flash welding and was given a quench and temper postweld heat treatment. Conditions of heat treatment, chemical composition, mechanical properties and dimensions are given in Table C-1. Heterogeneity in mechanical properties along the specimen may be estimated by the hardness distribution over a section. An example is given in Figure C-2.

Results of tensile tests are shown in Figures C-3, C-4, and C-5. Figure C-4 shows values of the ratio of the ultimate tensile strength of the joint, σ_u, and the ultimate tensile strength of the soft layer, σ_u^I, as a function of the ratio of the thickness of the soft zone, H_o, and the specimen diameter, D_o. The investigators called the ratio H_o/D_o the "relative thickness." Figure C-5 shows the ratio of the yield strength of the joint, σ_y, and the yield strength of the soft layer, σ_y^I, as a function of the relative thickness, X. The strength of the joint approaches that of the metal when X is sufficiently small.

*This literature review is a continuation of Chapter 8; the references cited herein are found at the end of Chapter 8.



Series	D_0 (mm)	H_0 (mm)	L_0 (mm)	X	Remarks (note)
I, II	10	025 – 190	80	003 – 190	
III	3	030 – 434	27	010 – 143	3A
	6	026 – 339	5.	021 – 156	3B
	10	060 – 139	80	006 – 138	3C
	15	104 – 152	135	007 – 101	3D

Figure C-1 Round-bar test specimen.

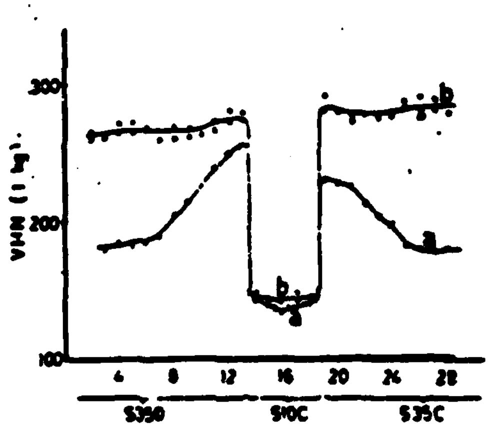
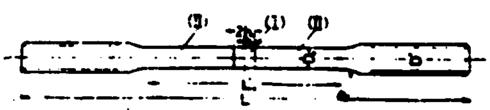


Figure C-2 Hardness distribution
over weld joint section.
(a) as weld,
(b) quenching (850°C)
and tempered (500°C).

TABLE C-1 Chemical Composition, Mechanical Properties and Dimensions of Specimens

Series	Test temp.		Material	Specimen diameter (D_0)	Content of elements, (%) C	Si	Mn	P	S	Mechanical particulars T.S. Kg/mm^2	Y.S. Kg/mm^2	R.A. %	El. %	n	HVD
I	room temp.	Soft interlayer	S10C	10	0.11	0.16	0.45	0.021	0.021	47.8	32.6	72.2	26.0	0.17	150
		Base metal	S35C	10	0.38	0.23	0.75	0.015	0.026	106.7	96.8	54.6	9.46	0.07	310
II	room temp.	Soft interlayer	S10C	3	0.11	0.16	0.45	0.021	0.021	47.1	31.5	76.3	27.0	0.17	150
				6						48.0	28.8	73.9	27.9	0.17	150
				10						47.8	32.6	72.2	26.0	0.17	150
		Base metal	S35C	3	0.38	0.27	0.72	0.027	0.021	86.1		72.1	33.3	0.07	280
				6						83.4	60.6	65.2	15.9	0.07	280
				10						83.3	64.1	67.7	14.5	0.07	280
	-73° C	Soft interlayer	S10C	10	0.11	0.16	0.45	0.021	0.021	56.6		69.3			
		Base metal	S35C	10	0.38	0.27	0.72	0.027	0.021	94.1	69.4	50.2			
III	room temp.	Soft interlayer	S10C	10	0.11	0.16	0.45	0.021	0.021	49.0	37.6	75.0	22.9	0.18	220
		Base metal	S30C	16	0.34	0.22	0.74	0.015	0.025	73.5	52.4	67.0	16.0	0.14	270

		Dimension (mm) D_0	D'	L_0'	L	R
Series II	Series A	3	5	27	80	10
	Series B	6	10	54	160	20
	Series C	10	14	90	210	20
	Series D	15	18	135	280	25



NOTE: T.S. : Tensile Strength
Y.S. : Yield Strength
n : Strain-hardening exponent
HVs : Hardness (Vickers Number)
R.A. ; Reduction in Area
El. ; Elongation in $4S_0$ mm

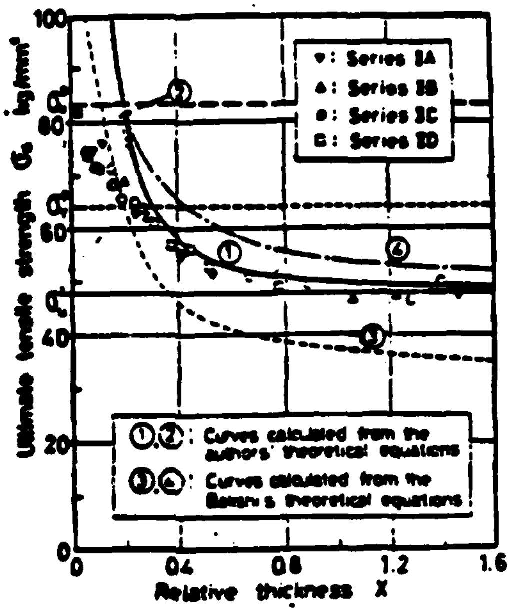


Figure C-3 Ultimate tensile strength σ_u as a function of relative thickness.

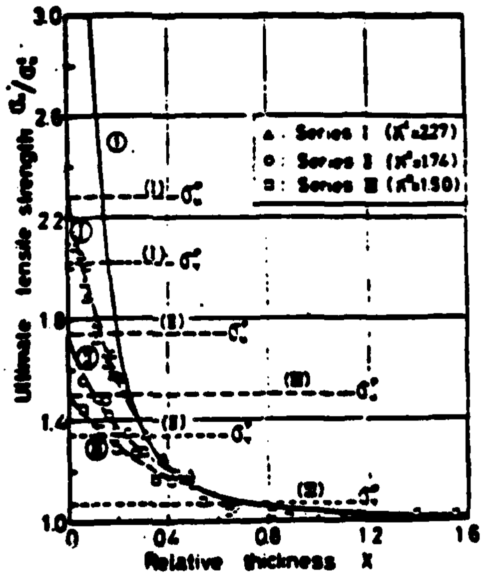


Figure C-4 Ultimate tensile strength σ_u/σ_u^I as a function of relative thickness.

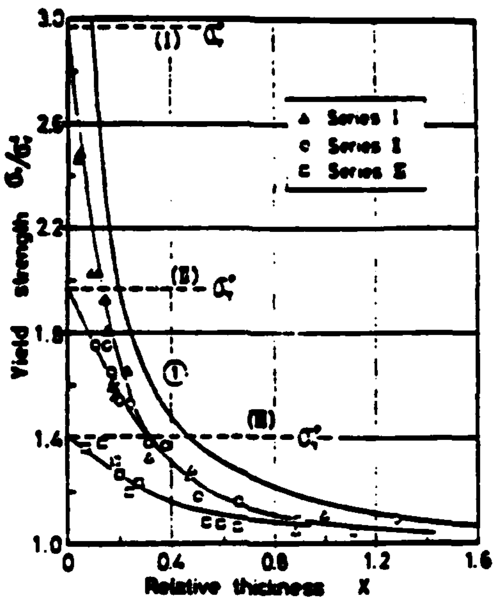


Figure C-5 Yield strength σ_y/σ_y I as a function of relative thickness.

Even though the maximum load was reached, appreciable reduction in diameter (necking) was observed at the soft interlayer where fractures occurred in all specimens. Reduction of area, ψ, (Figure C-6), decreases as X decreases to a certain minimum value, and then increases. For a sufficiently small value of H_o, the fracture mode changes from a fibrous cup-and-cone type to a brittle type with a rim of cleavage facets, even at temperatures at which the fracture mode of the virgin material is of the fibrous cup-and-cone type. The transition thickness for fracture location $(H_o)_{tr}$ depends on both testing temperature and base-interlayer strength difference (Figure C-7).

(B) Flat bar specimens, located across the weld line, were tested (Satoh and Toyoda 1970b). The specimens were prepared either by flash butt welding of round bars (19 mm ϕ; cross sectional area) of S15C and S35C structural steel (series A, B) or by shielded metal arc welding of 25 mm-thick plates of high-tensile steel HT-80 steel (minimum tensile strength 80 kg/mm^2), using electrodes capable of producing weld metal with a minimum tensile strength of 50 kg/mm^2 (series S, T). After welding, the specimens were given either quench and temper heat treatment (A, B) or stress relief treatments (S, T). Geometry and dimensions are given in Figure C-8, hardness distribution in Figure C-9, and chemical composition and mechanical properties in Table C-2.

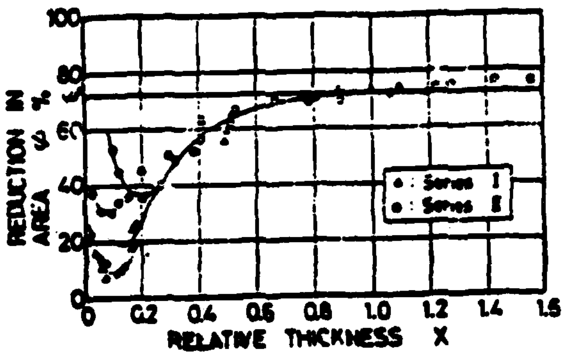


Figure C-6 Reduction of area as a function of relative thickness.

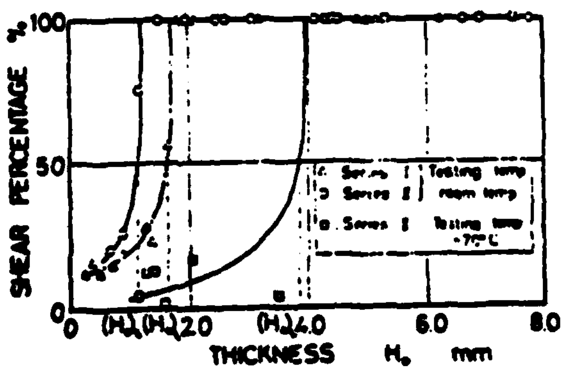
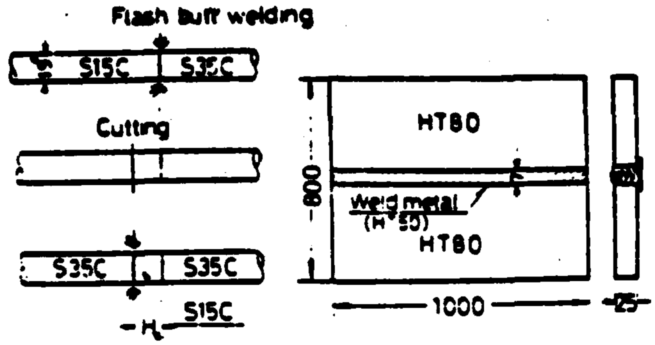
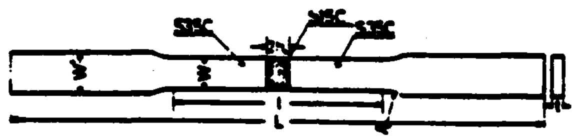


Figure C-7 Shear percentage as a function of thickness H_o (specimen diameter = 10mm).

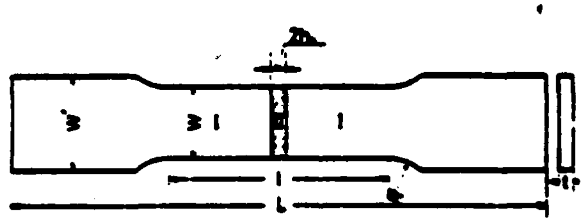


(a) Series A, B (b) Series S, T



	Dimension mm					
	l	L	W	W'	R	t
Series A	90	240	10	18	20	3
Series B	90	240	15	18	25	5

Dimensions of specimens (Series A, B)

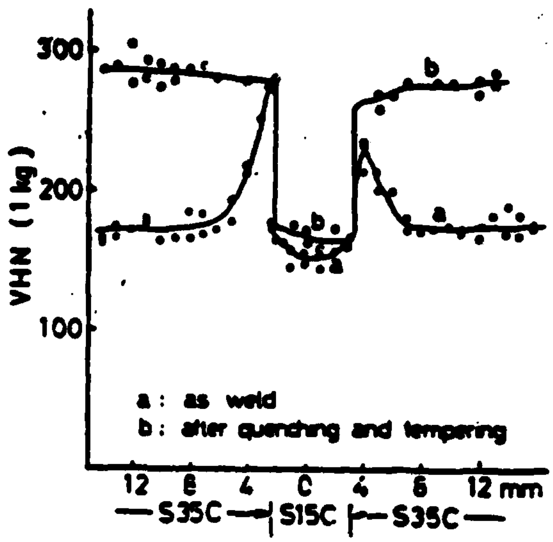


	Dimension mm						
	l	L	W	W'	R	t	2h
Series S	200	600	25~120	240~200	25~60	20	7
Series T	200	500	75~150	[illegible]~140	30~[illegible]	15	7

Dimensions of specimens (Series S, T)

Figure C-8 Dimensions of specimens used in some tests.

Series A & B



Hardness distributions over welded joint section (Series A,B)

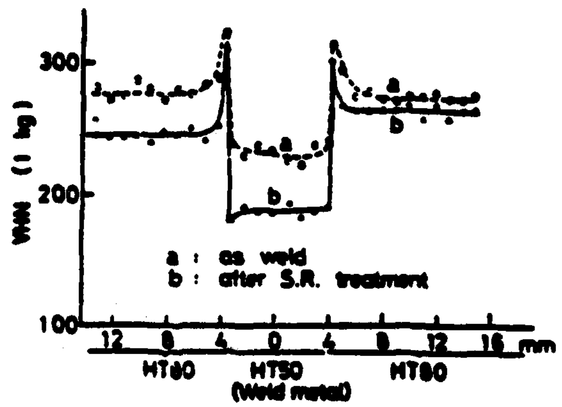


Figure C-9 Hardness distributions over welded joint section.

TABLE C-2 Chemical Compositions and Mechanical Properties of Test Specimens (Series A, B)

Chemical compositions and mechanical properties (Series A, B)

Material	Contents of Elements %					Mechanical Properties					
	C	Si	Mn	P	S	T.S. kg/mm²	Y.S. kg/mm²	R.A. %	El. %	n	VHN
S35C	0.37	0.26	0.63	0.017	0.020	84.0	64.5	55.9	15.0		280
S15C	0.11	0.22	0.50	0.021	0.015	47.8	34.4	76.8	27.9	0.18	165

(Specimen Diameter 10 mmϕ)

Chemical compositions and Mechanical Properties (Series S,T)

Chemical Composition

Material	Contents of Elements %									
	C	Si	Mn	P	S	Cu	Cr	Mo	Ni	V
HT 80	0.10	0.24	0.76	0.15	0.15	0.30	0.55	0.49	0.87	0.06
HT 50 (Weld metal)	0.09	0.62	1.03	0.013	0.006	0.04	0.03	0.01	0.02	0.03

Mechanical Properties

Material	T.S. kg/mm²	Y.S kg/mm²	R.A. %	El. %	n	VHN
HT 80 (base metal)	84.1	78.0	53.0	12.7	0.085	260
HT 50 (soft interlayer)	58.3	48.8	73.4	21.4	0.170	185

Note:
T.S.: Tensile Strength
El. : Elongation in 50 mm
R.A.: Reduction in Area
n : Strain hardening Exponent
Y.S. : Yield Strength
VHN: Vickers Hardness Number

Results of the tensile tests (Figures C-10 to C-12) indicate that the ultimate tensile strength and the yield strength of the specimen depend on both the relative thickness X_t, (H_o/t_o), and the ratio of plate thickness to width t_o/W_o. The joint strength increases as X_t decreases, reaching the strength of the base metal when X_t is sufficiently small. Figure C-10 suggests that as the t_o/W_o ratio increases, under a constant X_t value the ultimate tensile strength (U.T.S.) increases to a certain definite value that depends on X_t. The plate width, W_o, above which the strength becomes almost the same as that of an infinite plate is roughly five times the plate thickness, t_o, when X_t is less than one, and five times the interlayer thickness, H_o. This is also the case when the X_t ratio is larger than one (Figure C-13). The elongation is measured after fracture over a length $l_o = 2.26\sqrt{A_o}$ (A_o is the initial cross sectional area). The results are given in Figure C-14. Fracture occurred at the neck of the specimen and was of the cup-and-cone type for large values of X_t. As X_t decreases, however, the fibrous portion in the fracture facet increases (Satoh and Toyoda 1975).

(C) To simulate more applications, two series of experiments were carried out with loading parallel to the weld line (Satoh and Toyoda 1971). Specimen geometry of both series (A & B) is given in Figure C-15(a) and (b), with welding conditions in Table C-3, and hardness distribution along a specimen section in Figure C-16.

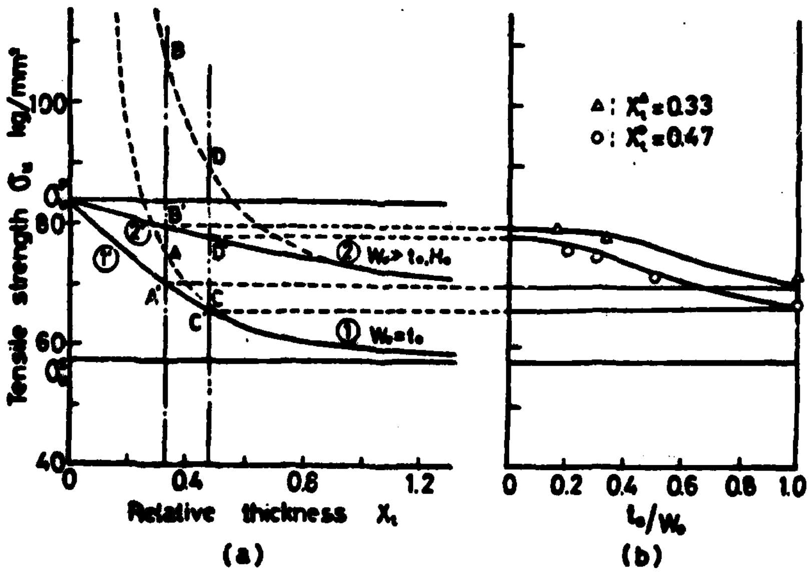


Figure C-10 Effect of plate width on ultimate tensile strength of welded plates (Series S,T).

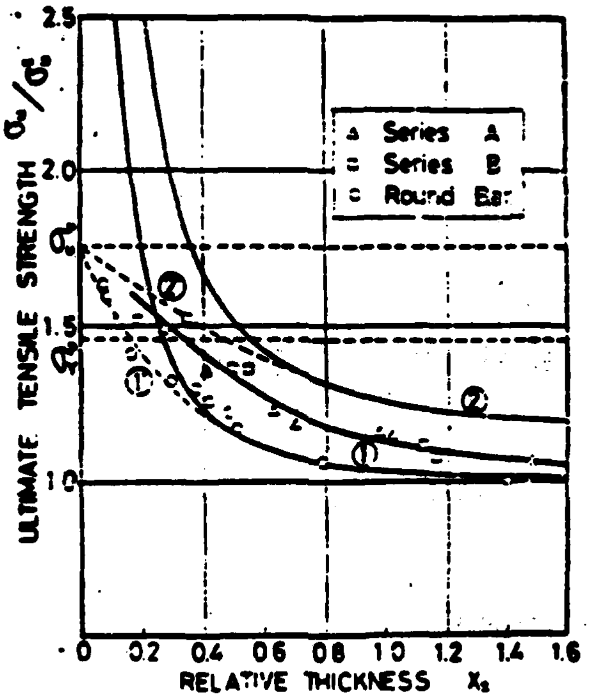


Figure C-11 Ultimate tensile strength as a function of relative thickness.

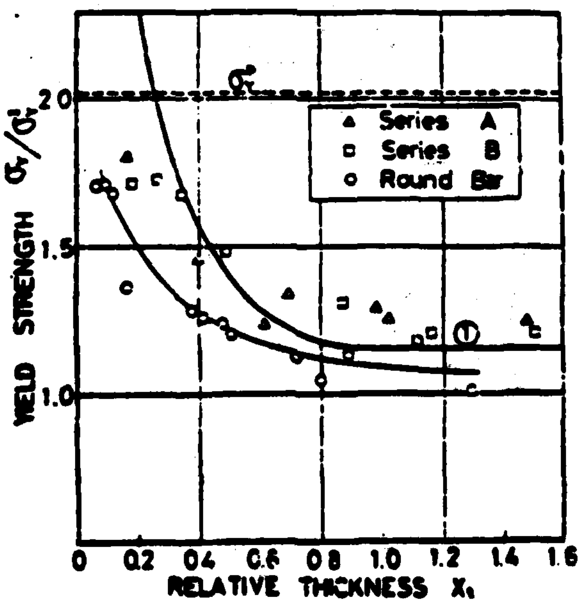


Figure C-12 Yield strength as a function of relative thickness.

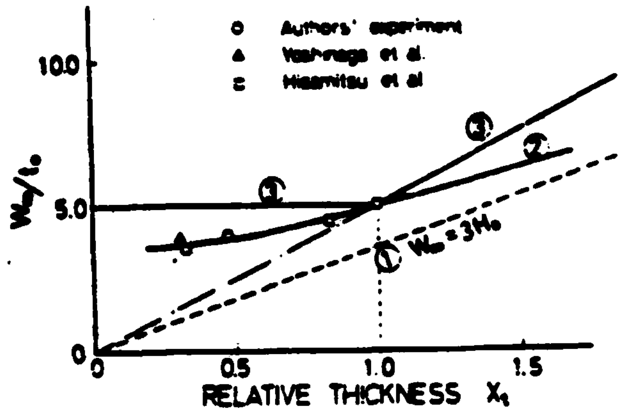


Figure C-13 The ratio W_{oo}/t_o as a function of relative thickness.

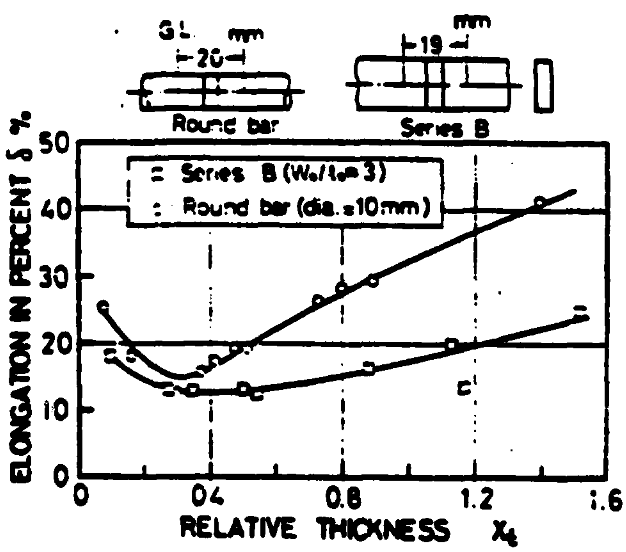
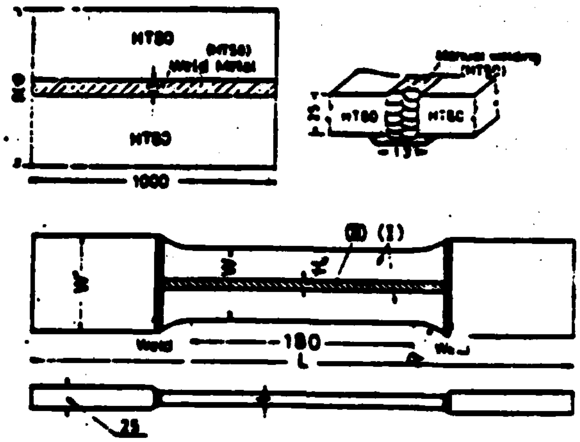


Figure C-14 Elongation in percent as a function of relative thickness (G.L. = 2.26 A_o).

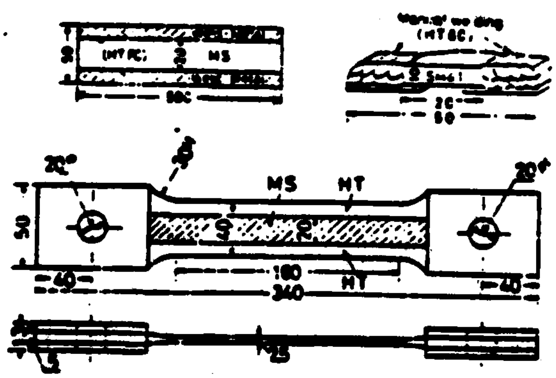


No.	Dimensions (mm)						m	γ
	W	t	W'	L	H_o	R		
A-30	30	15	50	460	13	25	1.3	0.57
A-60	60	15	75	530	13	40	3.6	0.78
A-100-1	100	15	125	640	13	50	6.7	0.87
A-100-2	100	5	125	640	13	50	6.7	0.87
A-150	150	15	185	850	13	55	10.4	0.91

(a) (Series A)

NOTE: $m = (W-H_o)/H_o$

$\gamma = (W-H_o)/W = m/(1+m)$



(b) (Series B)

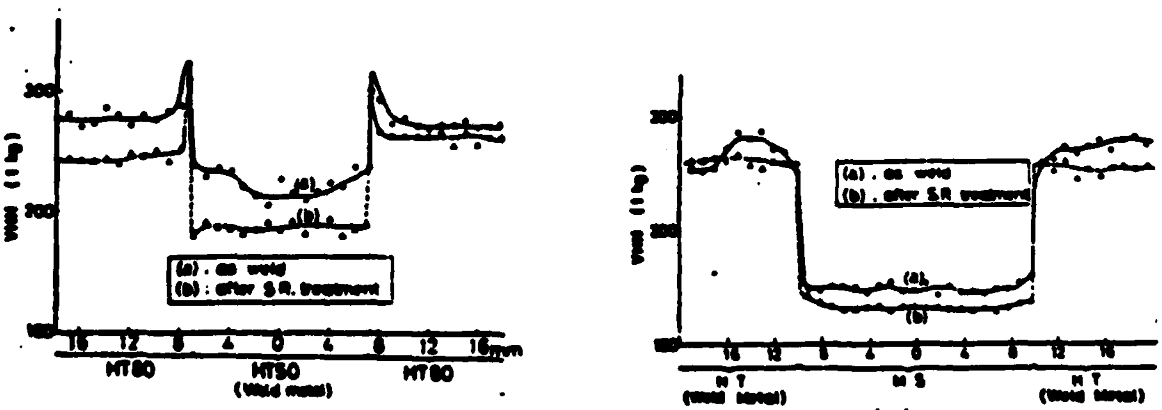
Figure C-15 Dimensions of specimens used in some tests.

TABLE C-3 Welding Conditions and Mechanical Properties

	Welding Conditions				
	Electrode diameter (mm)	Current (A)	Voltage (V)	Preheating temperature (°C)	
Series A	+	160	24	150	Low hydrogen type rod ($50\ kg/mm^2$)
Series B		170			Low hydrogen type rod ($80\ kg/mm^2$)

		Mechanical Properties			
		U.S. kg/mm^2	Y.S. kg/mm^2	R.A. °G	El. G.L. 80 mm
Series A	HT 80	84.1	78.0	53.0	12.7
	HT Weld M	58.3	48.8	73.4	21.4
Series B	HT Weld M	83.0	76.1	62.8	12.9
	MS (SM41)	41.5	22.3	59.6	26.1

Note: Specimen diameter 10 mm.



(a) (b)

Figure C-16 Hardness distributions over a specimen section.

The results of these experiments (Figures C-17, C-18) suggest that the strength and ductility of the joint depend on the ratio of the width of the hard zone to that of the soft zone (m-value) and become almost equal to those of the hard metal when m approaches 10 (see note in Figure C-15a). Strain distributions in the composite weldment are almost uniform along the mid cross-section of the specimen at each load except when yielding and after the maximum load (Figure C-19). Behavior of the axial strain around yielding seems to be influenced by the ratio of the width of the soft metal to the thickness of the plate. When this ratio is smaller than 2, the strain increases almost uniformly along a cross section until general yielding occurs. At that point, base metal strains are temporarily larger than those of the soft metal. When the ratio is larger than 2, non-uniform distribution of the strain occurs at an average stress somewhat larger than the yield strength of soft material and continues until general yielding.

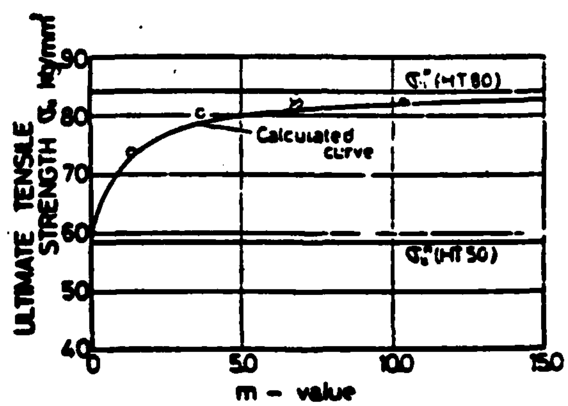


Figure C-17 Ultimate tensile strength σ_u as a function of m-value.

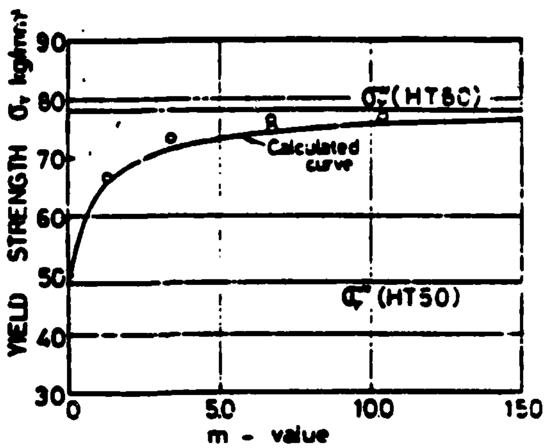


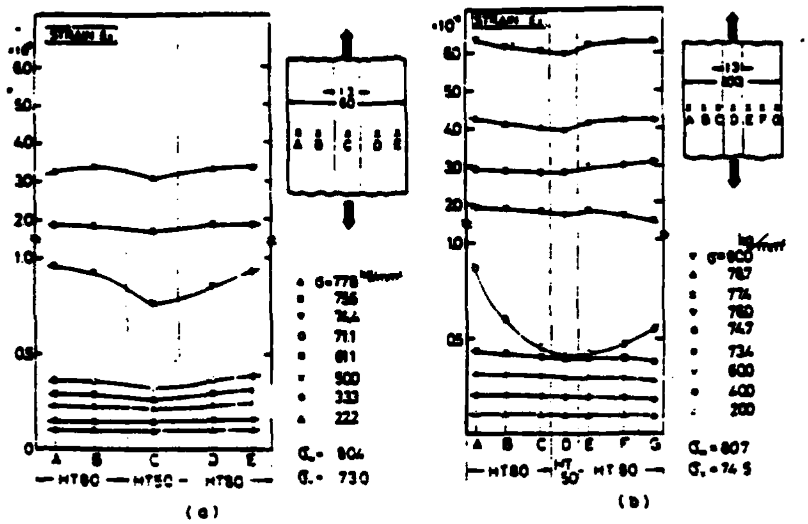
Figure C-18 Yield strength σ_y as a function of m-value.

The fracture mode is influenced by both the m-value and the thickness of the plate. As the m-value increases, the tendency toward oblique necking becomes pronounced after the maximum load is attained. The ratio of the width to plate thickness drastically affects the location and direction of macroscopic fracture.

Performance Study

To assess the applicability of undermatching joints in actual structures, the S. J. Committee of the Japan Welding Engineering Society carried out a performance study (Satoh and Toyoda 1971b, 1972; Satoh et al. 1979). Some of the results of the static tensile strength tests are summarized below.

Specimens (Figure C-20) were prepared by shielded metal arc welding or 70 mm thick HT-80 plate with preheat and interpass temperatures of 150°C and 250°C, respectively. The mechanical properties of the base metal and the electrode are shown in Table C-4. The weld metal shape is now realistic, and the average relative thickness $(X_t)_{av}$ is given by the formula in Figure C-20. Results summarized in Figure C-21 and C-22



Strain distributions along mid cross section for different stages of tensile load.

(a): Specimen A-60 (m = 3.6) (b): Specimen A-100-1 (m = 6.7)

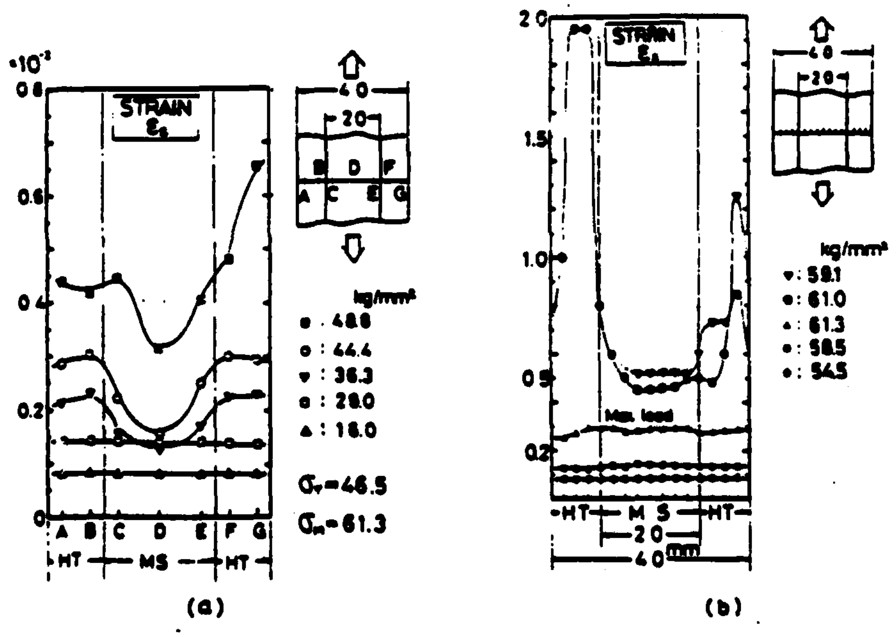


Figure C-19 Strain distributions along mid cross section of specimen (Series B).

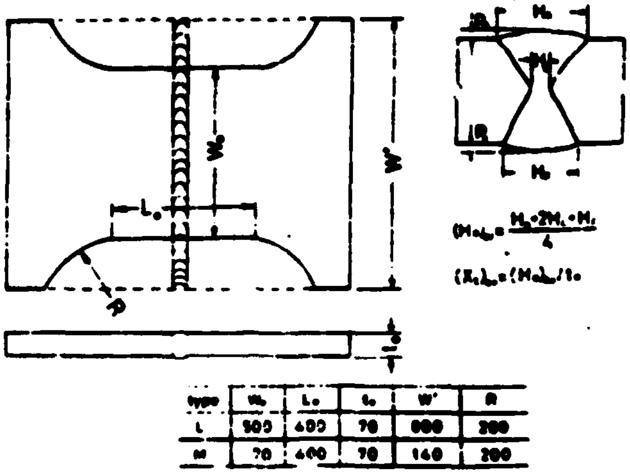


Figure C-20 Specimen design in static tension test.

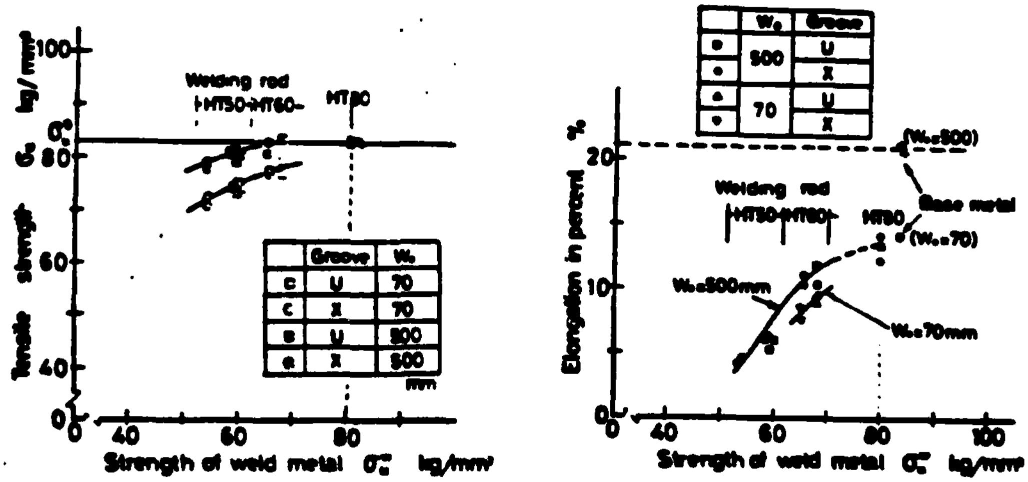


Figure C-21 Static tension test results.

TABLE C-4 Test Results of Base Metal and Weld Metal in Static Tension Test

Base Metal

Steel	Plate thickness	T S (kg/mm^2)	YS (kg/mm^2)	Elongation (percent)	Reduction in area (percent)
HT 80	70(mm)	821	754	254	592

(JIS No. 4)

Weld Metal

Specimen Number	Electrode	Groove	TS (kg/mm^2)	YS (kg/mm^2)	Elongation (percent)	Reduction in area (percent)
(1)-6U	E9016	U	678		24.5	67.5
(1)-6X	E9016	X	651		25.0	70.6
(1)-5U	E7016	U	596	514		75.2
(1)-5X	E7016	X	540	478		79.8
(1)-8	E11016		883	837		71.0

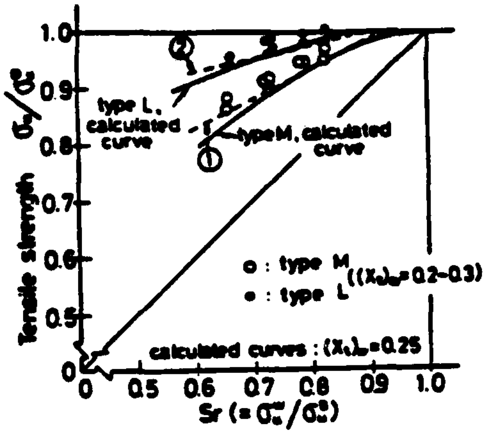


Figure C-22 Tensile strength ratio of undermatched welded joint versus undermatched weld metal. .

indicate that in square bar (M) specimens, the joint tensile strength (σ_u) does not reach the tensile strength (T.S.) of the base metal (σ_u^B), whereas in wide plate (L) specimens this happens even with the E9016 electrode. In all type-L specimens, fracture occurred very close to the weld fusion line and was accompanied by appreciable lateral contraction in the width direction. The joint ductility, however, becomes considerably smaller when using the E7015 (undermatched) electrode (Figure C-22).

As can be seen in Figure C-22 for the joint strength in L-type specimens, the value $(X_t)_{av}$ = 0.2 to 0.3 reaches the tensile strength of the base plate (σ_u^B) when the tensile strength of the weld metal (σ_u^W) is nearly 90 percent of ^{B}.

Tensile Strength of Undermatched Fillet Welds

The applicability of undermatching was also tested for fillet welds. Various specimen geometries were tested (Figure C-23) and some of the results are presented in Figure C-24. (Specimens were made of HT-80 steel.)

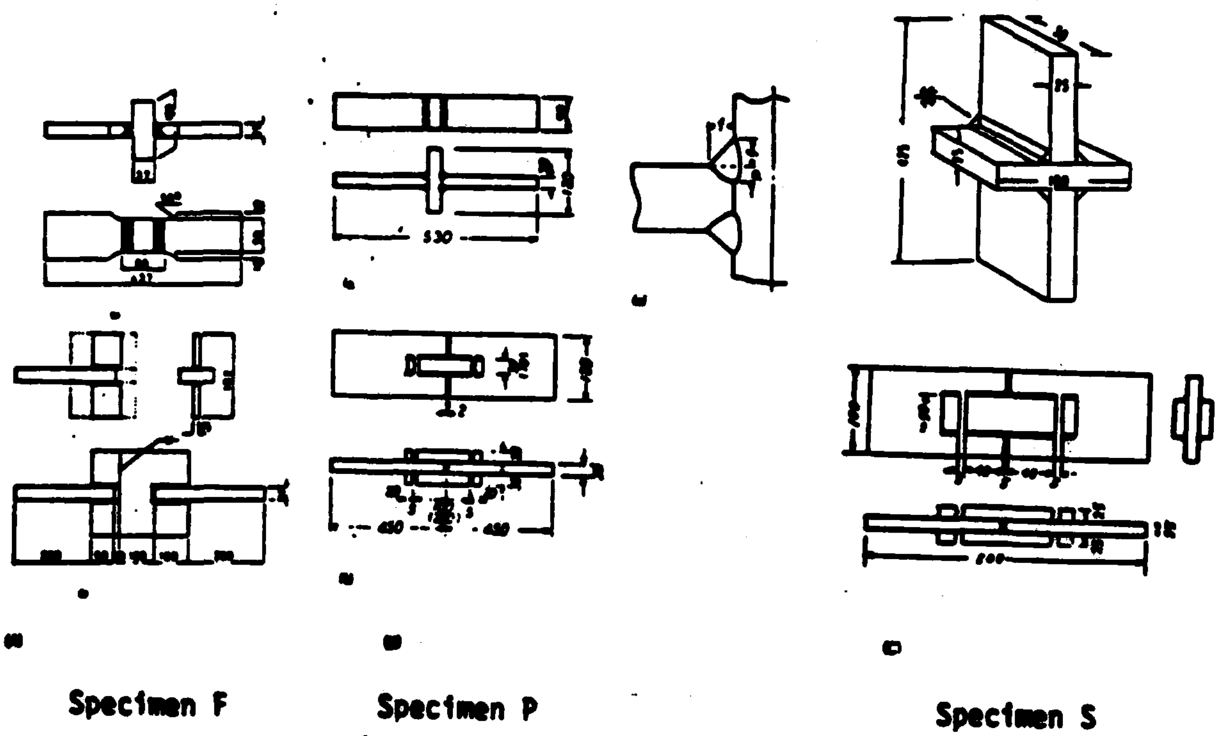


Figure C-23 Specimen geometry.

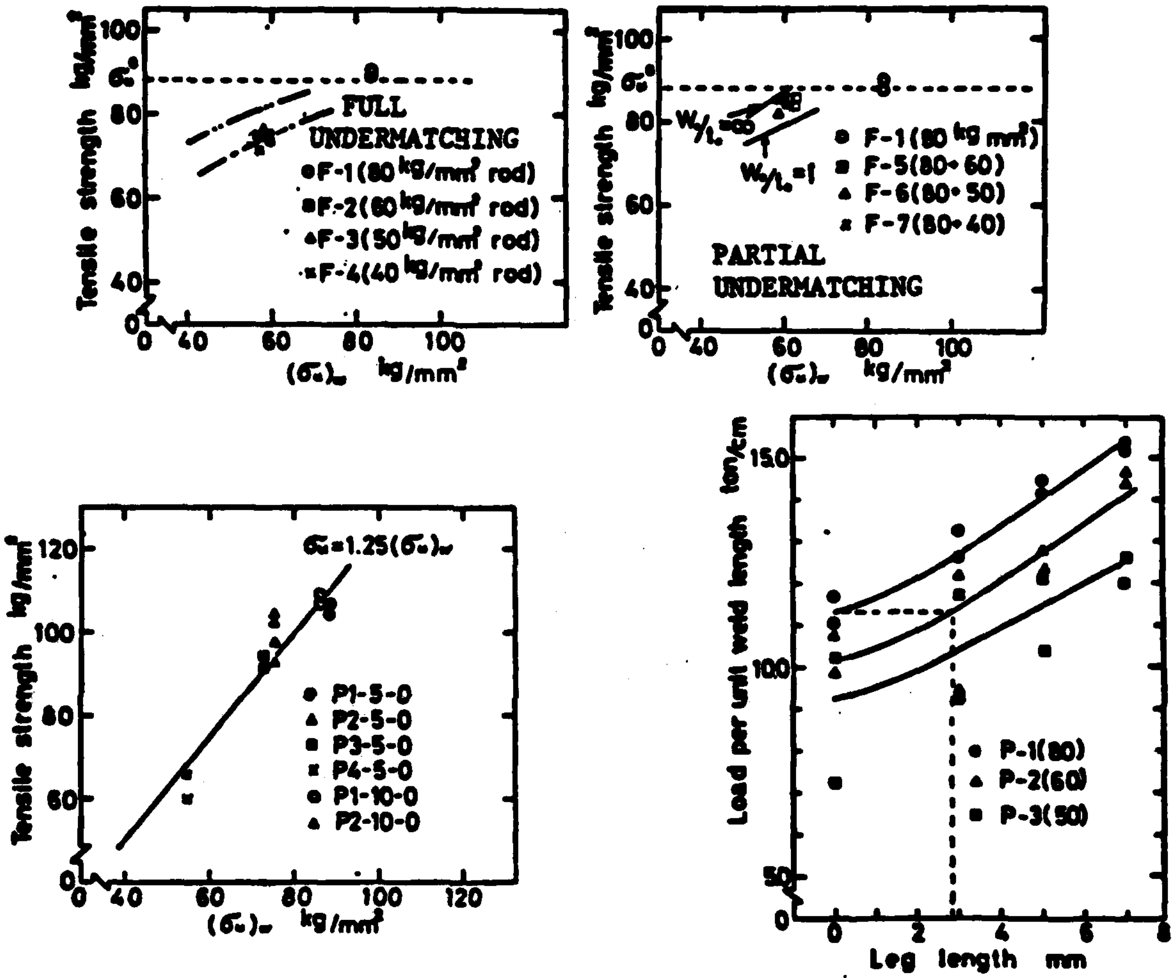


Figure C-24 Tensile test results.

Fatigue Strength of Undermatched Weld Joints

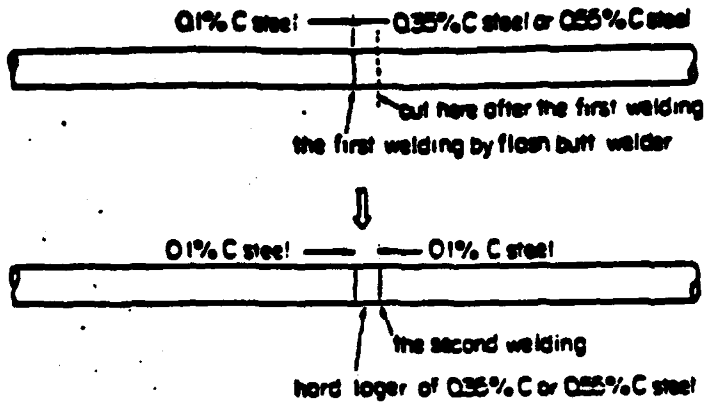
Satoh and Nagai (1969) performed tests to investigate the fatigue strength of both welded and locally work-hardened bars having hard or softer interlayers. Specimen geometries are shown in Figure C-25 and the hardness distributions resulting after the stress relief heat treatments are given in Figure C-26. Fatigue tests were performed with a rotating bending machine. The results shown in Figure C-27 indicate that the hard interlayer had no effect on the fatigue strength, whereas the fatigue limit decreases drastically when the thickness of the soft interlayer increases. Similar results were obtained with the locally hardened specimens.

To evaluate the performance of actual undermatched welded joints, the S. J. Committee of the Japan Welding Engineering Society conducted a series of fatigue tests using HT-80 specimens welded with E7016 and E11016 electrodes.

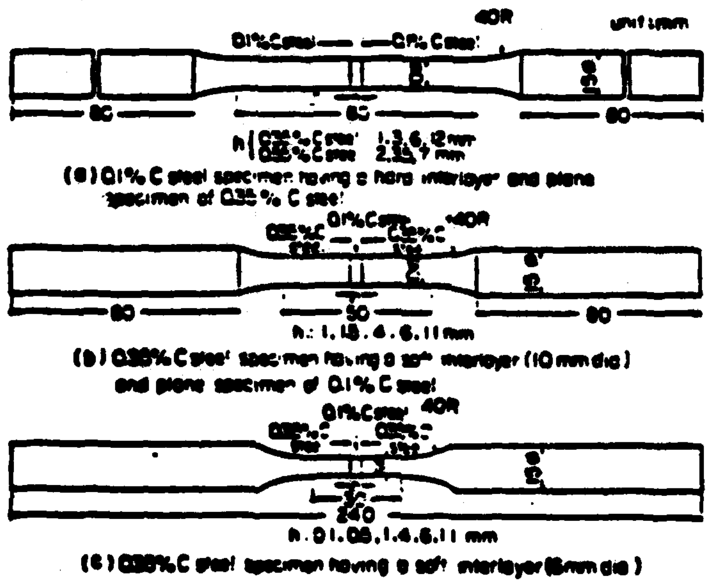
The first series of specimens tested (FT, FL) is shown in Figure C-28. The mechanical properties of the base metal and the weld metals together with fatigue test results are shown in Table C-5. The tests were under a pulsating load between zero and 35 kg/mm^2, and between zero and 55 kg/mm^2. In the FT specimens, where tensile stress is applied transversly to the weld, fatigue cracks occur at the toe of the reinforcement and propagate in the direction of the thickness. No appreciable difference in the number of cycles to fracture appears between the overmatched and the undermatched joints. In FL specimens, however, where tensile stresses are applied parallel to the weld, the fatigue life of the overmatched joint is somewhat longer because fatigue cracks are initiated on the surface of the weld metal. In both FL and FT specimens, the weld reinforcement had not been removed. Test results for other geometries are shown in Figure C-23. As indicated in Figure C-29c, the reinforcement decreases the fatigue strength of the welded joint. Work has also been completed on the fatigue strength of undermatching fillet welds by both the S.J. Committee and other investigators.

Fracture Strength of Undermatched Welded Joints

Satoh and Toyoda (1973) directed an extensive investigation into the brittle fracture strength of undermatched weld joints. Two series of experiments were conducted with deep-notched specimens (Figure C-30). The specimens of HT-80 plates were prepared by submerged arc welding using electrodes having a nominal tensile strength of 50 to 80 kg/mm^2. The tests were performed between 0 and -200°C (Figure C-31). The σ_y curves refer to the 0.2 percent proof stress of each material. The fracture stress curve of the joint for the BI-type tests moves to the higher temperature side as compared to the curves of the base metal. Joints welded with 60 kg/mm^2 electrodes, which happened to have the lowest fracture transition temperature, seem to have somewhat higher strength. In the BII-type tests, the residual stresses caused by welding were considered responsible for the observed two-step fracture. These tests indicate that a lower transition temperature should be required for the undermatched weld metal than that for the overmatched one. This was also confirmed by the theoretical analysis discussed in Chapter 8.



(A) Preparation of welded specimens (D=19mm)



(B) Geometry of fatigue bars prepared from welded specimens.

Figure C-25 Shape of welded specimens.

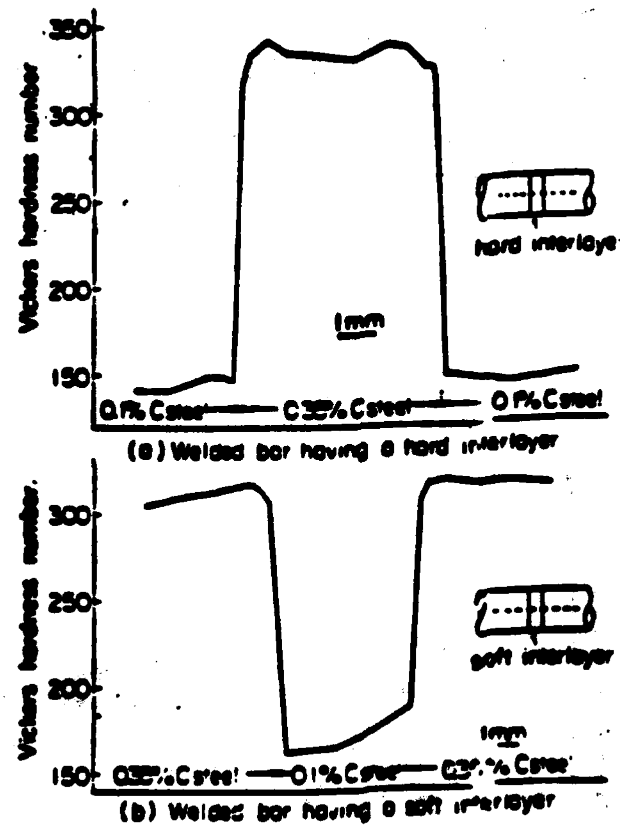
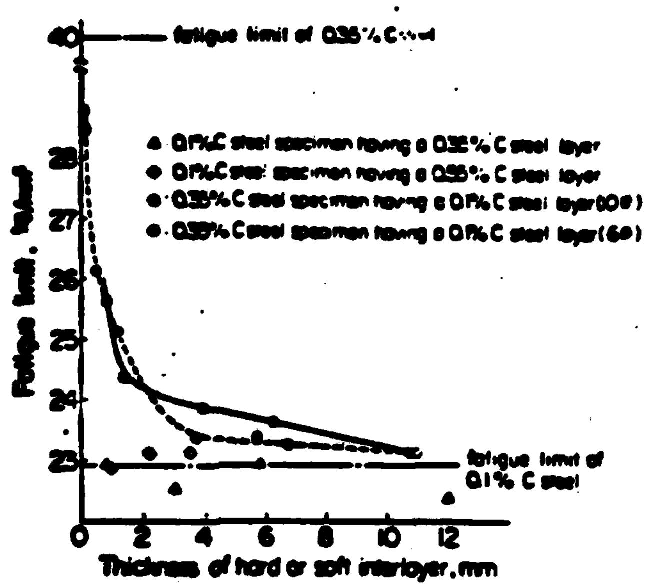
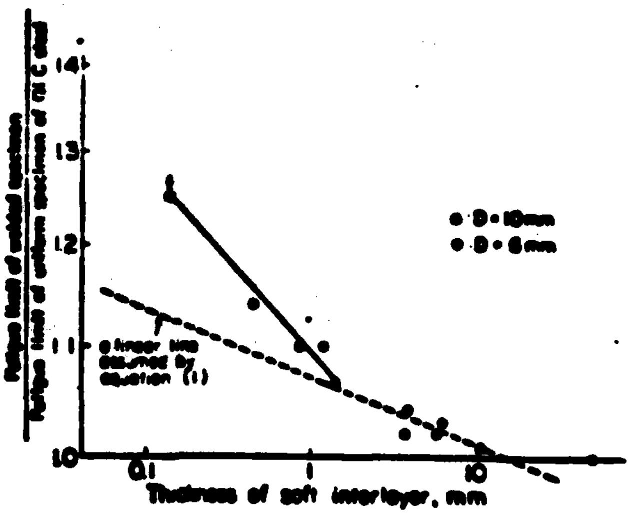


Figure C-26 Hardness distribution in the welded bars.



(a) Fatigue limits of welded bars having a hard or soft interlayer.



(b) Relation between normalized fatigue limit and thickness of a soft interlayer.

Figure C-27 Fatigue test results.

TABLE C-5 Mechanical Properties of the Base and Weld Metals

Material		Tensile Strength (Kg/mm^2)	Yield Strength (kg/mm^2)
Base Metal HTBO		82	79
Weld	E 11016	86	74
Metal	E 7016	55	-

(a) Chemical composition: C 0.10%, Mn 0.79%, Si 0.26%, P 0.004%, S 0.007% Ni 0.83%, Cr 0.52%, Mo 0.34%

Maximum stress applied (kg/mm^2)	Electrode used	No. of cycle at fracture Type FT ($\times 10^2$)	Type FL ($\times 10^2$)
35	E 11016-G	15.2	20.8
35		13.2	23.1
	E 7016	13.8	16.2
		14.1	22.3
		3.72	2.62
	E 11016-G	1.38	2.45
		3.17	
55		8.65	4.63
	E 7016	6.63	5.34

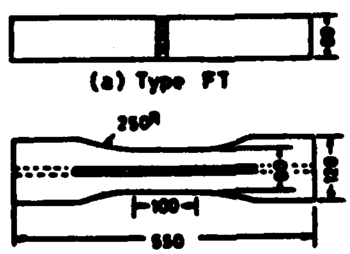


Figure C-28 Type FT and FL specimens.

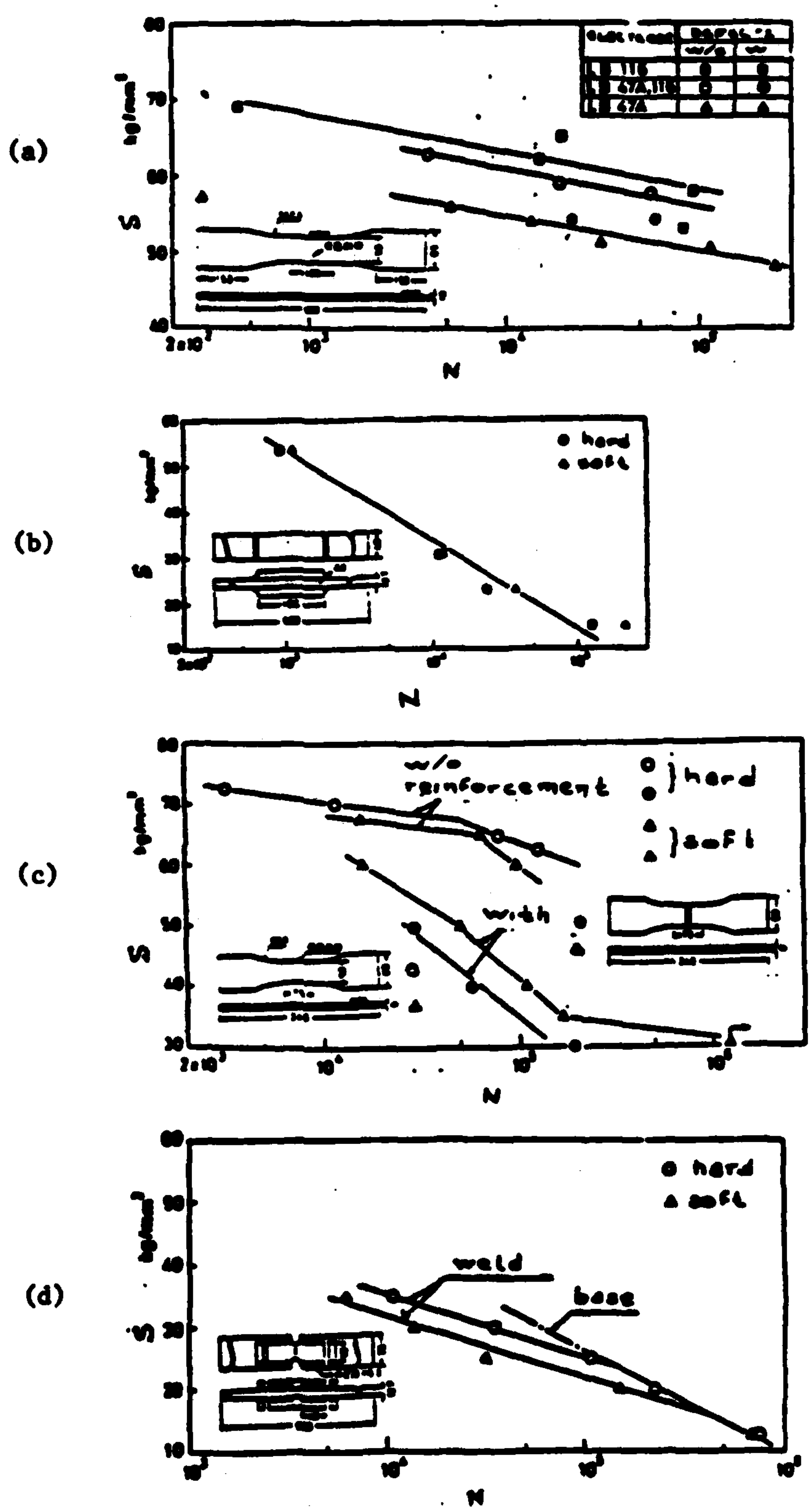


Figure C-29 Fatigue test results.

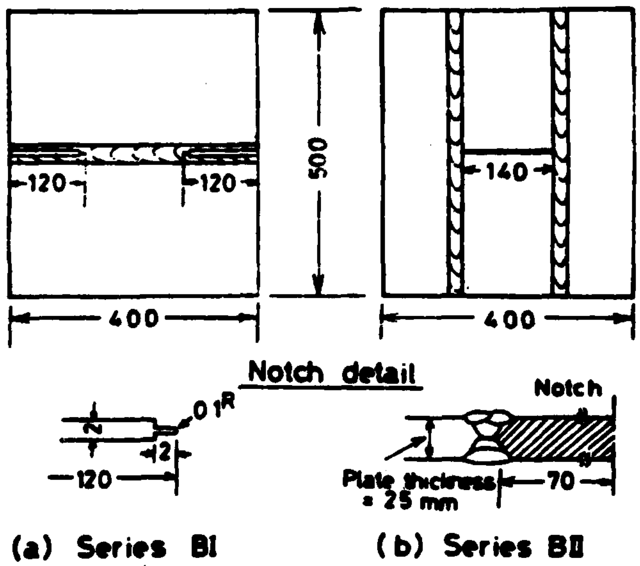
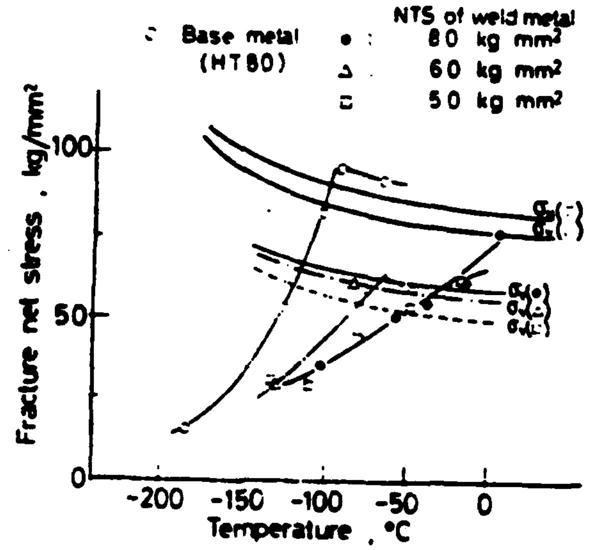
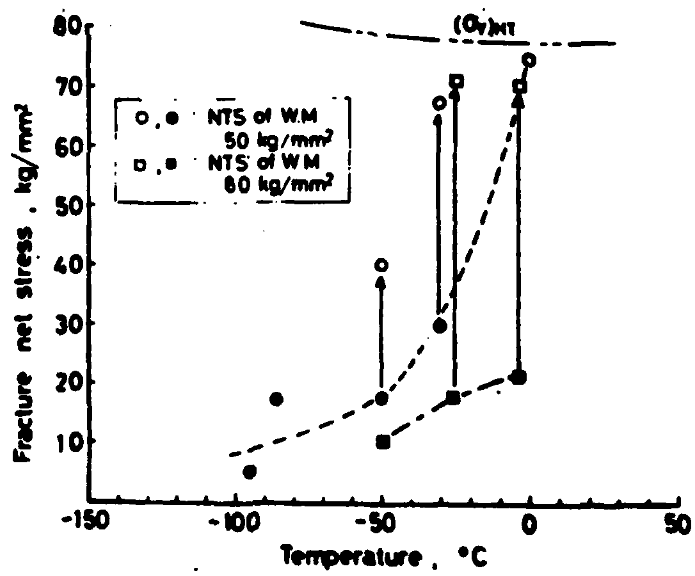


Figure C-30 Brittle fracture test specimens.



(a) Fracture stress versus temperature relations in BI-type test.



(b) Fracture stress versus temperature relations in BII-type test.

Figure C-31 Brittle fracture test results.

APPENDIX D

NAVAL RESEARCH LABORATORY
WASHINGTON, D.C. 20390

6381-23M:PPP:svh
NRL Prob No. M01-25
SER:

2 MAR 1971

From: Director, Naval Research Laboratory
Washington, D. C. 20390
To: Commander, Naval Ship Engineering Center
(D. E. Smith, Code 6101D)

Subj: NAVSHIPSYSCOM 0900-005-5000, "Standard for Preproduction Testing Materials by the Explosion Bulge Test," revision and updating of

Encl: (1) Comments on revision and updating of NAVSHIPSYSCOM 0900-005-5000

1. The Naval Ship Engineering Center, Code 6101D (Mr. D. E. Smith), requested that the Naval Research Laboratory, Code 6380, review and comment on the subject report.

2. The review was completed and comments are forwarded as enclosure (1).

W. S. PELLINI
By direction

Copy to:
NAVSHIPSYSCOM (Code 2052)
NAVSHIPSYSCOM (Code 031)
NAVSHIPSYSCOM (Code 03423)
NAVSHIPRANDLAB (Code A833)

NAVSHIPSYSCOM 0900-005-5000, "Standard for Preproduction Testing Materials by the Explosion Bulge Test," revisions and updating of P. P. Puzak (Note: Underlined items denote revisions by Puzak.)

The following comments concern the revisions to NAVSHIPSYSCOM 0900-005-5000 that were listed in the Naval Ship Research and Development Laboratory Report NP/3960 (A933 RS), Work Unit 1-933-503, Report 9-37, of 5 January 1971. The revisions in these documents were considered as appropriate except for those sections listed below. Comments on the proposed revisions are mainly in the form of a discussion to justify the need for modification in the proposed revisions.

Section 5:5.3.2, page 6, paragraph 1, line 14: recommended change from "0.707 inches to 1/16 of an inch": The origin of this change is typographical in nature and should be corrected to read "0.079 inches (2 mm)

Section 6:6.13, page 8, paragraph 1, lines 1, 2, 3, 5, 6 and 7 - This section concerns the use of the Charpy V-notch (C_v) test to establish the upper shelf energy level and the effect of the temperature range from room temperatare to -60°F on C_v energy. Recent reviews of the effectiveness of the C_v test to define accurately the temperature transition features of steels have emphasized the problems of interpretations of C_v energy in the transistion region (1). The intent of the C_v test in NAVSHIPSYSCOM 0900-005-5000 was to ascertain whether the steels subjected to this test were operating at their upper shelf level of fracture resistance at 0°F. It is now possible to obtain a more quantitative measure of this critical parameter for HY-130 (or higher strength) steel products than that provided by the C_v test. Extensive documentation for the justification of the Dynamic Tear (DT) test for this purpose is provided in references (2-10). To accomplish the objectives of the explosion bulge test, definitive knowledge of fracture resistance at the minimum service temperature of 0°F is an adequate criterion for acceptable service performance. This information can be obtained with duplicate 5/8 in. DT tests, and modification to the revision of 6:6.1.3 is recommended as follows:

6:6.1.3 Dynamic Tear (DT) Test. Two each top and bottom 5/8 DT specimens shall be taken from the weld metal as per Fig. 10 and paragraph 7.5.2 of NRL Report 7159, "Standard Method for the 5/8 Inch Dynamic Tear Test." The specimens shall be notched perpendicular to the surface. Light chemical etching of the specimens is recommended in order to locate the notch wholly within and on the centerline of the weld metal. The four 5/8 DT specimens shall be tested at 0°F (-18°C) and the energy absorption results recorded.

Section 6:6.2 and Section 7:7.2. The revisions recommended for these sections are not adequate to preclude unwarranted crack starter or explosion bulge tests of materials performing below minimum acceptability levels of performance. Also, as presently written the subject document permits a waste of time and money for unnecessary explosive shots on specimens for which

testing should be terminated. It is recommended that sections 6:6.6 and 7:7.2 of the subject document be modified extensively and for purposes of clarity and continuity, the revised sections with NRL suggested changes underlined below are as follows:

6.2.1 Two weldments modified with crack starters shall be tested prior to the explosion bulge tests. These specimens are used for quick screening, to determine whether a continuation by explosion bulge testing is warranted. The purpose is to develop an early crack which results in the catastrophic propagation of a fracture if the material, weld, heat-affected zone, or fusion line have tendencies for low-energy propagation of the crack.

6.2.2 Only materials and processes which perform above minimum acceptability levels of performance in the crack-starter test as defined in section 7.2 shall be further evaluated by explosion bulge tests. Explosion bulge tests require the application of repeated explosure shots to delineate the critical regions of the weldment which participate in the initiation and propagation of fracture. Acceptable performance must be exhibited by a minimum of four explosion bulge specimens. The specimens shall be refrigerated to a temperature sufficiently below the test temperature so that heat gain during handling will not cause the test temperature to be exceeded. Test temperature shall be as specified by the Naval Ship Engineering Center. The refrigerated specimens shall be placed on the die with the ground ends of the weld in contact with the die. Using the stand-off and charge specified, one shot shall be fired. Stand-off distance shall be measured from the under side of the plate (from the face of the die) to the bottom surface of charge. The charge shall be carefully centered over the center of the die and the blasting cap shall be inserted in order to direct the explosion downward. The fired specimen shall be carefully examined and the location, length, and direction of all cracks recorded (preferably by a sketch or photograph), together with a description of the fracture appearances, if visible. In the absence of cracks or separations in the explosion bulge test, the maximum reduction in plate thickness at a point near the apex of the bulge (approximately 1-1/2 inches from the edge of the weld) shall be measured after each shot and recorded as a percentage of the original plate thickness.

6.2.3 In the explosion bulge testing of HY-80 material, the development of less than 3 percent thickness reduction as a result of the first shot shall be considered inadequate. The charge weight and stand-off distance shall be modified for explosion bulge tests of the remaining specimens of that group until 3 percent minimum thickness reduction per shot is attained. In the explosion bulge testing of HY-130 material, the development of less than 2.5 percent thickness reduction as a result of the first shot shall be considered inadequate. The charge weight and stand-off distance shall be modified for the testing of the remaining specimens of that group until 2.5 percent minimum thickness reduction per shot is attained. Stand-off distances less than 15 inches shall not be employed for any test. Before each successive shot, the specimen shall be returned to the cold box sufficiently long (about 3 hours) to equalize the required temperature throughout the specimen. Succeeding shots shall be fired using the same stand-off and charge and the results shall be recorded.

6.2.4 In the event a visible failure is developed in a specimen, the fracture surface shall be examined, and location, length, and direction of failures shall be recorded as above-together with the percentage reduction in specimen thickness developed on the shot preceding the failure shot. The specimen shall be returned to the cold box sufficiently long to equalize the test temperature throughout the specimen. Using the same stand-off and charge, another shot shall be fired. Thickness reduction measurements shall not be taken.

Section 7 - Acceptance Rejection Standards

7.1 Mechanical Tests

When base metals are being evaluated, the base metal shall meet the requirements of the procurement specification. The deposited weld metal shall be proven acceptable by meeting the requirements of the appropriate weld metal procurement and fabrication specifications. When filler metal or welding processes are being evaluated, the deposited weld metal shall be proven sound by nondestructive testing and proven acceptable to appropriate weld standards. Further, the mechanical properties of the base metal and weld metal shall be proven to meet the requirements of their respective procurement specification.

7.2 Explosion Crack-Starter Tests

7.2.1 Materials or procedures being investigated which develop the following indications of high brittleness after one shot in the crack-starter explosion test shall be rejected without conducting the "next" shot or any explosion bulge tests of the remaining specimens.

7.2.1.a Development of fractures extending into or through the hold-down regions of the specimen.

7.2.1.b Development of transverse (square) weld-cracks which penetrate the weld thickness.

7.2.2 The extension of fractures after two shots into or through one or both ends (hold-down regions) of the crack starter test specimen causing complete separation of one or both weldments shall be considered to be below minimum acceptability levels, and explosion bulge testing of the remaining specimens shall not be undertaken.

7.2.3 The development of short tears after the first shot, with only moderate extensions (to but not through the hold down region) of these fractures by the second shot, shall be considered to be above the minimum acceptability levels of performance in the crack-starter explosion test. Providing both crack-starter specimens indicate performance above minimum acceptability levels, explosion bulge testing of the remaining specimens shall be warranted.

7.2.4 Upon receipt of two explosive loadings the crack-starter specimens shall exhibit no cracking outside the bulged region. When testing plate properties, fracture through the weld metal shall be considered no test. When testing weld metals, fracture through the parent plate or heat-affected zone shall also be considered no test.

7.3 Explosion Bulge Tests

7.3.1 The development of a full hemispherical bulge without fracture is a function of the relative strain hardening characteristics and flow strengths of the weld deposit, heat-affected zone and plate areas near the apex of the bulge. The maximum reduction in thickness of the plate at a point near the apex (approximately 1-1/2 inch from the edge of the weld) shall be measured after each shot and recorded as a percentage of the original plate thickness. The percent thickness reduction of plate shall not of itself constitute a definition of full hemispherical bulge without fracture, because this parameter will vary with the geometry of the die, the flow strengths of various areas in the weldment and the thickness of the sample being tested. Testing of any given sample shall be discountinued upon attainment of full hemispherical bulge with no visible evidence of fractures.

7.3.2. Unless otherwise directed by Commander, Naval Ship Engineering Center, Hyattsville, Maryland, explosion bulge tests shall be terminated after sufficient explosive loadings which result in 12 percent thickness reduction of the plate for HY-80 materials and 10 percent thickness reduction of plate for HY-130 materials. When testing plate properties, failure through the weld metal shall be considered no test. When testing weld metal properties, failures through the plate shall be considered no test.

7.3.3. Materials or procedures being investigated which develop the following type first visible fractures in explosion bulge tests shall be rejected without continuing the "next" shot.

7.3.3.a The development of first visible fractures which consist of transverse (square) weld-cracks which penetrate the weld thickness.

7.3.3.b Transverse or longitudinal fracture which extend into or through either end (hold-down region) of the specimen.

7.3.3.c For specimens given a next shot, extensive propagation of first visible failure indications into or through either end (hold-down regions) of the specimen shall also be considered below the minimum acceptable levels of performance in the explosion bulge test.

Enclosure (1) to NAVSHIPRANDLAB ltr Ser MP/3960(A933 RS) Work Unit 1-933-503 Report 9-37 - Revised Figure 4. - Delete weld metal Charpy V specimens and substitute two each (top and bottom) 5/8x1-1/2x7 in. DT specimens.

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