A REVIEW OF WELDING CAST STEELS AND ITS EFFECTS ON FATIGUE AND TOUGHNESS PROPERTIES

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OUTLINE OF THE PROBLEM

Welding is extensively employed in the finishing stage of steel casting production and in fabricating larger components by joining castings or by joining castings to wrought steel components. The ease of welding as well as the effects of welding on mechanical properties are thus very important considerations in the production and selection of castings as engineering components. The purpose of this review is therefore to summarize and evaluate the published literature with respect to weldability of carbon and low alloy steels and the effects of welding processes on mechanical properties.

SUMMARY OF CONCLUSIONS

The technical literature has been reviewed and analyzed to present the significant factors that determine the weldability of cast steels and the effect of these welds on the fatigue and toughness behavior. The effect of welding processes, the significant variables of these processes, the chemical composition, pre- and post-welding treatments and presence of discontinuities on the mechanical properties are indicated with emphasis on the fatigue strength and toughness,

The weldability of carbon steels decreases with increasing carbon, alloy and sulfur contents. The higher carbon and alloy steels require higher temperature preheats and greater use of postheating treatments to obtain satisfactory welds. Fatigue properties are improved by removal of weld reinforcements, the use of low hydrogen shielding, automatic compared to manual welding, full heat treatments and peening. The toughness of welds is optimized by: low carbon, oxygen, nitrogen, sulfur and phosphorus contents with a minimum alloy content for the required strength level; low heat inputs and multipass welds; low hydrogen arc shielding and basic fluxes; and flat welding positions and automatic welding processes.

The presence of discontinuities can lower both the fatigue and toughness properties markedly. The effect of these discontinuities varies with the stress system under dynamic loading. Generally, the most damaging discontinuities listed in order of decreasing severity are: cracks, undercuts, slag inclusions, porosity and the presence of weld reinforcements.

PREFACE TO 1979 PUBLICATION

This publication "A Review of Welding Cast Steels and Its Effects On Fatigue and Toughness Properties" has been issued based on a report issued in 1974 to members of the Steel Founders' Society of America. This publication has been prepared as a supplement to the new report "Repair Welding and Fabrication Welding of Steel Castings" issued by the Steel Founders' Society in 1979.

PREFACE TO 1974 REPORT

Advances in welding technology, in quality control of welds, and in performance of welded structures have led to increased customer acceptance of welding as a regular part of the foundry production process. Welding is now recognized as a procedure for upgrading casting quality during the course of manufacture through improvement of surface conditions, or by elimination of shrinkage voids. It is also accepted as a method of producing large or complex assemblies where the size of the completed structure precludes production as a one-piece casting, or where total quality will be improved by dividing the structure into simpler components which can later be welded into an integral assembly.

With the customer's and casting user's acceptance of the welding process, welding has gained a position where the foundryman has to be cognizant of all the welding processes, their capabilities, precautions necessary for satisfactory weldability, quality control and the effects of welding on properties of the weld and the weld affected areas of the casting. Steel foundries have recognized these implications quite early. A booklet, "Recommended Practice for Repair Welding and Fabrication Welding of Steel Castings", was therefore first published by the Steel Founders' Society of America in 1957, with a revised and updated version following in 1969.

The above publication on welding practices has been of great benefit to steel foundries and their customers. Member foundries, however, have shown increased interest in the question of weldability which means the ease of welding or the precautions to be taken to assure a satisfactory weld, depending on the welding process and alloy to be welded, as well as the effects of welding on properties and performance of the weld and weld-affected area.

The Carbon and Low Alloy Technical Research Committee therefore recommended an in-depth survey and evaluation of the published literature. This recommendation was approved by the Board of Directors in 1973 as Research Project 95.

Professor Wallace, at Case Western Reserve University in Cleveland, was asked to undertake this activity. The Research Committee compliments his effort that yielded a unique report where the important aspects have been assembled and reviewed with outstanding competence and insight. Acknowledgements are also made to Walter E. Evans, Ralph D. Maier and John C. Rogers, graduate students at Case Western Reserve University who assisted in the sections on weldability, fatigue strength, and toughness portions, respectively.

PETER F. WIESER
Technical and Research Director
By direction of the
Technical Research Committee

A REVIEW OF WELDING CAST STEELS AND ITS EFFECTS ON FATIGUE AND TOUGHNESS PROPERTIES

I. INTRODUCTION

This report reviews and analyzes the American and British literature on the weldability of carbon and alloy steel castings for total alloy contents up to 5%. The influence of various welding processes and techniques including electrode selection, pre- and post-treatments on the susceptibility to welding discontinuities and the mechanical properties of the welded structures has been studied. Considerable emphasis has been placed on the dynamic properties, fatigue and toughness because of the significance of these on service behavior. This report first discusses the weldability of the various types of steel and the problems with welding discontinuities and then the effect of the welding processes and welding discontinuities on the mechanical properties.

The weldability of steel castings is of considerable significance because the welding process is employed extensively for the repair of discontinuities and for fabricating larger components from castings. Repair welding and fabrication practice for steel castings was the subject of an earlier publication of the Steel Founders' Society of America (1). The use of welding procedures to join steel castings to other castings or wrought steel to produce larger structures is an established commercial procedure (2).

II. WELDABILITY

No specific criterion of weldability has been generally accepted; the term is used to describe the ease with which a metal can be welded to produce a weldment of acceptable quality. The weld quality is usually judged from the standpoint of mechanical properties (3):

(1) The strength of the joint must be at least as great as that of the parent metal;

(2) The fracture ductility of the weld metal and heat affected zone (HAZ) must be sufficient to ensure that the brittle fracture properties of the structure in service are not limited by these factors alone;

(3) The fatigue properties of the joint should not be impaired by the metallurgical condition of the weld metal or HAZ;

(4) The metallurgical condition of the joint should not impair the behavior of the structure during service as a result of localized corrosion, etc.

The metal which can be welded to fit the above criteria with no special precautions to prevent discontinuities or other difficulties is considered to have good weldability.

The strength, ductility, and to a large extent, the susceptibility of the weldment to cold cracking can be controlled by ensuring that the weld metal and HAZ have the proper metallurgical structures. These structures are determined mostly by the chemical composition and cooling conditions. It is therefore desirable to provide close control of both variables. The most common and effective means of controlling cooling rates in the weld area is to preheat the casting to be repair welded or fabricated to increase the base metal temperature and to decrease thermal gradients around the weld. This preheating may be localized around the weld area by using gas torches and various types of insulation. However, a general preheat of the entire casting is preferred when ecomomics and the welding operation permit. This general preheat can also be conducted in various types of portable equipment composed of gas torches and insulating or protective hoods. Again, it is preferable to preheat in a permanent recirculating gas or oil fired oven where the castings are placed in the oven and brought to temperature before their removal. In any case, some means should be available of insuring that the castings are up to the required preheating temperature before welding is initiated and maintained at that temperature throughout the welding operation. Surface reading thermocouples or crayons that melt at a specific temperature are commonly used for this purpose.

The specific preheat temperature is dependent on the steel composition, section size, the degree of restraint at the joint, the welding method and the ambient temperature. Preheating temperatures may vary from 100°F to over 1000°F, but most recommended temperatures are 600°F or less (4). The benefits obtained from proper preheating include: the prevention of cold cracks; the reduction of HAZ hardness; and residual stresses and distortion. The recommended preheating temperatures are discussed for each of the various types of steel in the subsequent sections.

Types of Discontinuities

One of the considerations of weldability is freedom from weld discontinuities. Just as the casting process may result in discontinuities that are frequently repaired by welding, the fusion welding process is susceptible to certain irregularities that may require correction for some applications. These weld discontinuities are generally repaired by additional welding (5). Some types of welding processes and steels are more susceptible than others to these discontinuities; the weldability is considered to be better for these steels and methods where less difficulities are encountered.

The major types of weld discontinuities in fusion welds are listed below and are shown schematically in Figure 1 (5).

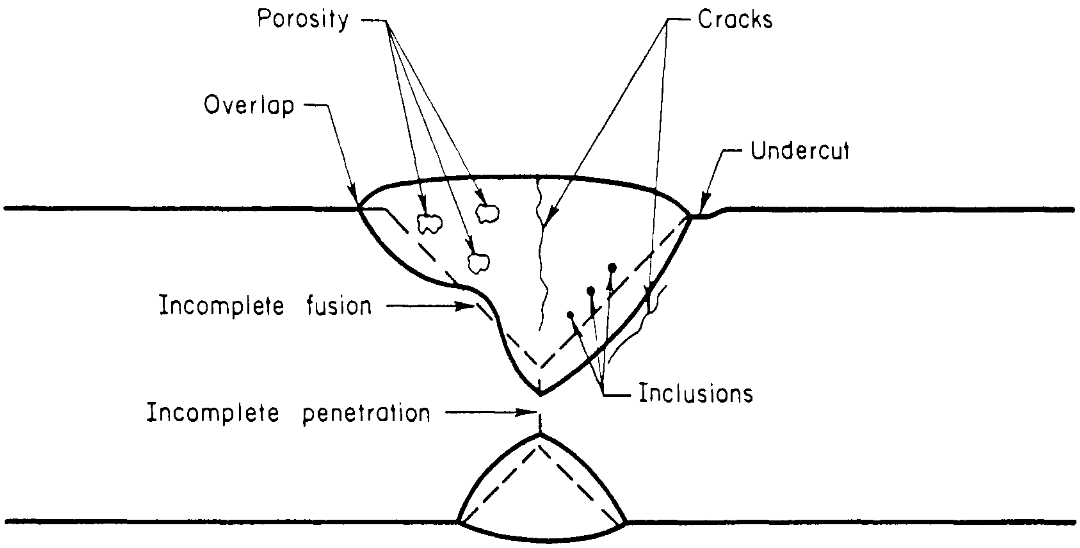


Figure 1. Major fusion weld discontinuities (5).

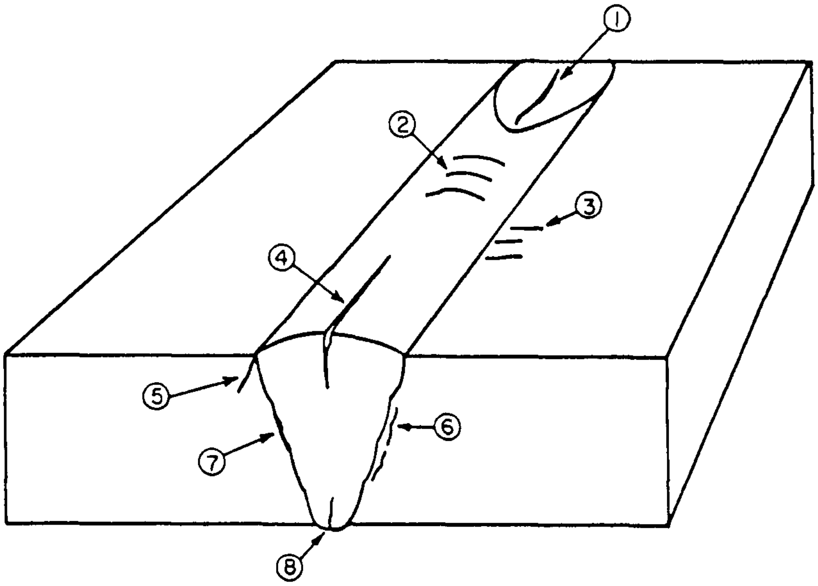


Figure 2. Weld cracks classed according to location (4).

1. Weld metal crater cracking
2. Weld metal transverse cracking
3. Base metal heat-affected zone transverse cracking
4. Weld metal longitudinal cracking
5. Toe cracking
6. Underbead cracking
7. Fusion line cracking
8. Weld metal root cracking

1. Incomplete fusion and joint penetration.

2. Inclusions: oxides, slag and tungsten.

3. Geometric imperfections: undercutting, underfill, excessive reinforcement, surface irregularities, dropthrough and mismatch.

4. Metallurgical Defects:
   a) Defects related to segregation:

      hot cracking and microfissures;
      cold cracking, delayed cracking, porosity
      and subsurface shrinkage.

   b) Imperfections induced by metallurgical reactions:

      embrittlement;
      metallurgical notches.

5. Other imperfections: arc strikes, weld spatter (5).

Incomplete fusion discontinuities are conditions in a welded joint where adjacent layers of weld metal, base metal or weld metal and base metal fail to melt together properly. They usually result from: lack of penetration of the full base metal thickness; or beads fail to intermelt; or the presence of slag, oxide or other foreign materials at interfaces prevent melting of adjacent materials. Incomplete-fusion discontinuities may be the result of faulty welding techniques and improper welding conditions. Insufficient welding current or voltage, or excessive root thickness is a frequent cause. Misalignment of the welding torch and the line of welding can also produce them. Incomplete-fusion with previous beads often results when deep crevices or undercuts are present in the previously deposited beads or when slag removal is incomplete. Although incomplete-fusion discontinuities are common in welds, they can be eliminated entirely by the use of proper welding conditions (5). Incomplete penetration both reduces the load bearing cross section and acts as a stress concentrator.

Inclusions in welds are undesirable foreign matter that was introduced during the welding operation. Oxide films, slag from the welding electrodes, and tungsten particles from gas tungsten-arc welding electrodes are the inclusions usually found in fusion welds. These oxide films can occur with improper shielding during welding. Slag inclusions are solid non-metallic particles that usually occur as continuous or intermittent stringers held between weld beads, or between weld beads and the base metal. Slag discontinuities occur mainly in welds made with the shielded metal-arc and submerged-arc processes that utilize a welding flux. Inadequate slag-removal techniques and/or improper welding techniques cause them. Tungsten inclusions are particles of tungsten in weld metal deposited with the gas tungsten-arc welding process and are condidered to have the same effect as slag inclusions of equal size. Like incomplete fusion, inclusions can reduce weldment properties and service performance, depending on their characteristics, orientation and the service conditions (5).

Geometric imperfections are weld-shape features such as undercut, excessive reinforcement, poor profile, excessive weld width, and others that result from improper contouring of the weld. They usually result from improper welding conditions or inadequate control of welding operations. Undercut denotes depressions along the edges of a weld, and parallel to its length. They form as intermittent or continuous grooves which vary in depth and width. Undercut occurs when surface tension forces draw the metal away from the sides of the weld groove. The weld reinforcement is that amount of deposited weld metal that projects beyond the surface of the base metal in the thickness direction. It can result from high heat input rates. The most common irregularities in the surface of welds are surface ripples. Although they are difficult to avoid, ripples can generally be removed by grinding or machining (5).

Cracks are a very serious form of welding discontinuity and are classified as hot cracking, cold cracking and fissuring. Hot cracks are essentially small hot tears and formed for the same reasons. Hot cracking at the weld centerline is encountered with weld pools of a tear drop shape and avoided by elliptical shape weld pools. Cold cracks occur at or near room temperature and can form a considerable period after the weld has been made. Cracks are also classified according to location as illustrated in Figure 2 (6). This type of classification is related to the conditions producing these cracks and cracking problems are frequently referred to by these designations in this paper. Underbead cracks are cold cracks occurring in the HAZ under the deposited weld and are produced by excessive hardness and brittleness in the HAZ, stresses in the area or hydrogen. Toe cracks have similar causes but can be observed at the surface of the weldment (5).

Porosity consists of numerous gas pockets in the weld resulting from the evolution and entrapment of gas from molten metal during solidification. Although the pockets usually are spherical in shape with bright, smooth walls, they vary in size and shape. Porosity also may be present as irregularly shaped gas pockets along dendrite boundaries, or as tubular gas pockets; "pipe" or "wormhole" porosity. The gases that cause porosity may come from welding materials themselves, from welding unclean or wet components, or from the use of improper welding conditions (5).

Metallurgical Discontinuities

An understanding of metallurgical discontinuities or structural variations in the welding of cast steel requires a knowledge of the transformations that occur with the heating and cooling that accompanies fusion welding. Steel is basically an iron-carbon alloy so the effect of the various thermal cycles on the phases obtained in the iron-carbon system is an important consideration. The influence that these thermal fluctuations exert on these phases during welding of a steel containing 0.3% carbon is illustrated by the simplified iron-iron carbide diagram in Figure 3 (6). Five points or locations have been selected to illustrate the various structures and conditions that occur as a result of the maximum temperature attained at each location.

As shown in Figure 3, some of the base metal in the HAZ (Points 1-4) is heated into the austenitic or high temperature (face-centered cubic) form of iron by the fusion welding operation. As these areas cool down to room temperature, this austenite transforms back to the room temperature phases with their body-centered structures. This transformation can occur to form several microstructural forms such as ferrite, pearlite, bainite or martensite. These constituents have different characteristics and their hardness generally increases for the phases formed at the lowest temperatures. Martensite, the phase that is formed at the lowest temperature, is both very hard and brittle at any except the lowest carbon contents. For this reason, the formation of martensite in the HAZ can lead to cold cracks, hydrogen-induced cracks or high residual stresses either during or subsequent to welding. Since the formation of such cracks and stresses reduces the mechanical properties, welding conditions are controlled for the different steels produced to avoid the formation of this martensite. Those steels that require special precautions to avoid martensite forming in the HAZ are considered to have inferior weldability to those steels that do not require special handling.

The carbon content, critical cooling rate and critical temperatures as affected by composition and the grain size of the steel are the primary factors that determine the tendency of each steel fo form martensite. Increasing the carbon content decreases the critical cooling rate, lowers the A_3 temperature and increases the hardness of the transformed region (4). The A_1 and A_3 temperature may be predicted with fair accuracy from the alloy content such as by the expressions listed below that are useful for low-alloy steels containing up to 0.60% carbon :

$$A_1(°F) = 1333 - 25(\%Mn) + 40(\%Si) - 26(\%Ni) + 42(\%Cr)$$

$$A_3(°F) = 1570 - 323(\%C) - 25(\%Mn) + 80(\%Si) - 32(\%Ni) - 3(\%Cr).$$

The higher the steel is heated above the A_3 temperature, the larger the austenitic grain size becomes; this effect is significant in the highest temperature region of the HAZ (Point 1 in Figure 3) (4).

On cooling from the austenitic region, each steel has a critical cooling rate though the region in which austenite transforms to the lower temperature microconstituents. Since the equilibrium diagram is not very useful in predicting structures obtained under rapid cooling conditions, continuous cooling diagrams (CCT) have been constructed for a large number of steels. These diagrams allow a correlation between cooling rates and the resulting microconstituents formed. The cooling rate is determined by welding conditions; slow cooling rates are favored by high heat input from the welding operation, thinner sections and high preheating temperatures. A CCT diagram for a low alloy steel (AISI 8630) is shown in Figure 4 (5).

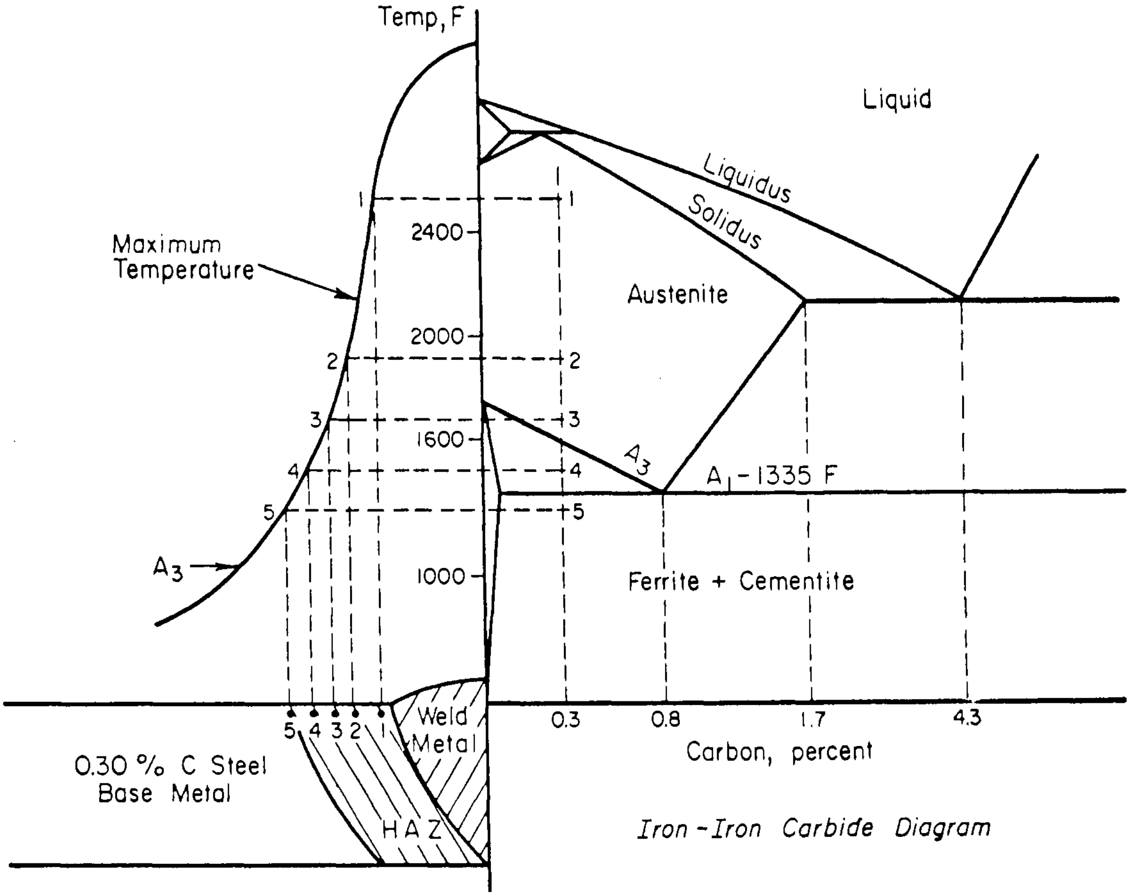


Figure 3. Base-metal peak temperature/weld-heat-affected zone iron-carbon equilibrium diagram correlation (6).

Point 1 has been heated in excess of 2400°F. The austenite that forms will be coarse-grained because of the grain growth at this temperature

Point 2 has been heated to 1800°F and fully austenitized. Grain growth is not a problem.

Point 3 has been heated to just above the A_3 critical temperature which is not high enough to provide completely homogeneous austenite during the welding cycle.

Point 4 This area has been heated to approximately 1400°F, which is between the A_1 and A_3 critical temperature shown in Figure 3. Part of the structure is converted to austenite and the resulting mixture of products during cooling can result in poor notch toughness.

Point 5 This point has been heated to 1200°F, which is below the lowest critical temperature, and no austenite has formed. Instead, the base metal may be softened by tempering.

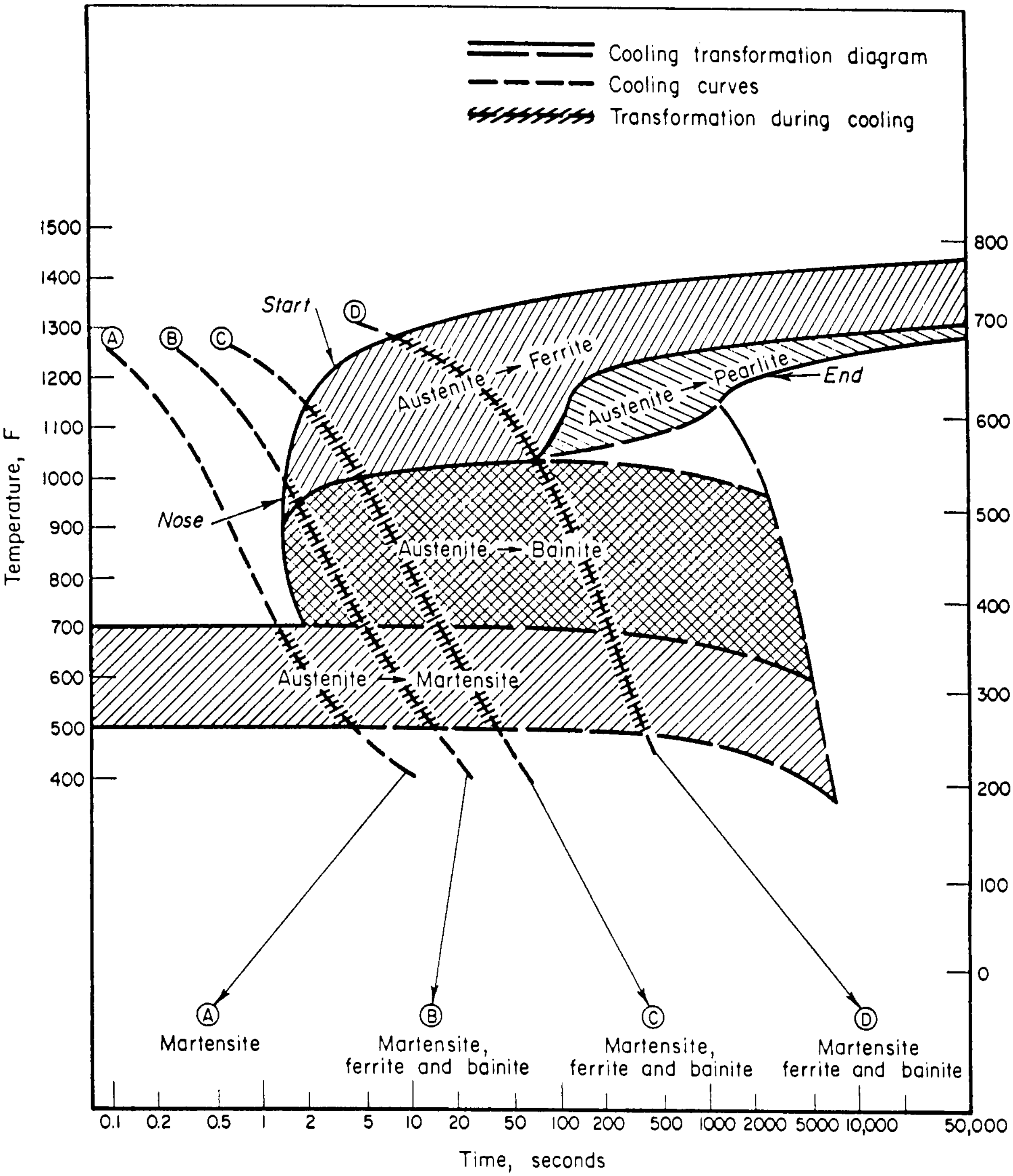


Figure 4. Transformation diagrams showing microstructural dependence on cooling rate (AISI 8630 Steel) (5).

It becomes evident from a study of this diagram that the different cooling rates shown by curves A, B, C and D provide different final microstructures with different hardness levels. The critical cooling rate as described above is defined as the maximum rate that will just provide a completely martensitic structure. This critical rate is located just to the right of curve A in Figure 4. From a welding standpoint, partially hardened or partially martensitic structures are undesirable even though small amounts of martensite can be tolerated in some cases.

The phases obtained during transformation of the austenite at various coolingratesor the CCT diagram for each steel varies with alloy and carbon content. These CCT diagrams are available for a large number of steels and allow this correlation between the cooling rate and structure. Steels with carbon contents under 0.10% have extremely rapid critical cooling rates and therefore present no problem with hardening even though the M_s temperature increases with decreasing carbon.

For a given steel, the final structure can be reasonably approximated by means of microhardness tests taken in the HAZ if correlating tables are available. A large amount of work has been done to relate welding variables, cooling rates and hardness obtained for various steels. Data are available to show the relation between heat input, cooling time through the transformation temperature region and section size for various steels and welding processes (7).

Welding Processes

The primary welding processes that are used for the repair and fabrication of steel castings are listed below and illustrated by the sketches in Figure 5 (8).

A. Shielded Metal-Arc Welding (SMAW) involves the use of a consumable electrode shielded by a gas originating from the electrode coating or in some cases, the inner core of the electrode.

B. Gas Metal Arc Welding (GMAW) , also known as Metal-Inert Gas Welding (MIG), the consumable electrode is shielded by an inert gas from a second source; such gases are helium, argon (sometimes mixed with small amounts of hydrogen) and CO_2 for certain steels,

C. Gas Tungsten Arc Welding (GTAW), also known as Tungsten-Insert Gas Welding (TIG), a gas shielded tungsten electrode is used to deposit metal from a consumable welding rod.

D. The Submerged Arc Welding (SAW) process has an arc maintained between a continuously fed consumable bare wire electrode and the work. Flux is dispensed over the area to be welded and the wire is fed into and melts some of this flux.

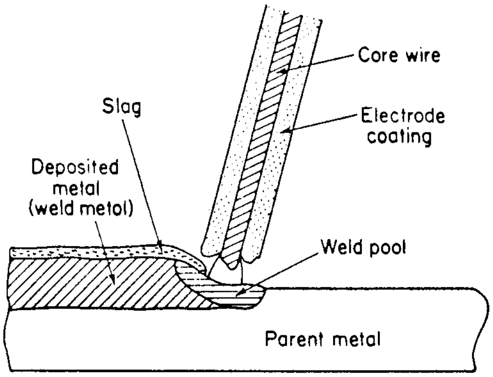


Figure 5a. Arc welding with coated electrodes (8).

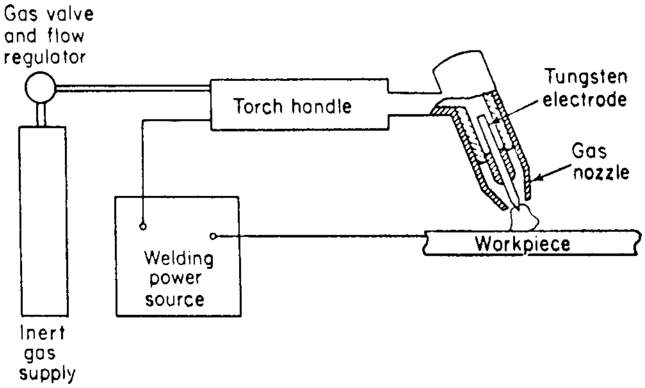


Figure 5c. Tungsten Inert gas welding equipment (8) (GTAW).

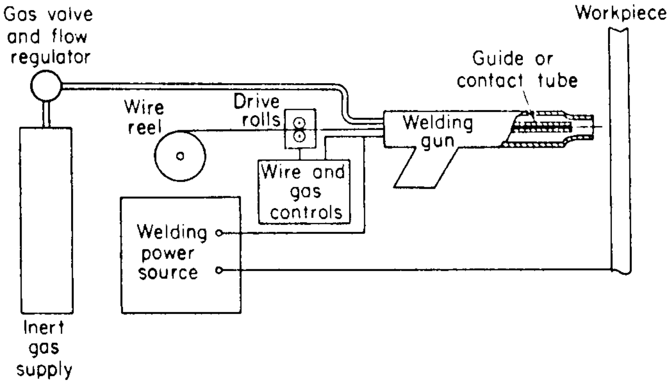


Figure 5b. Metal inert gas welding equipment (GMAW) (8).

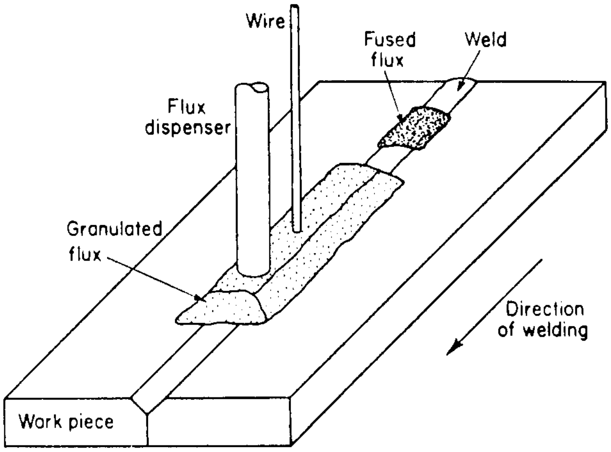


Figure 5d. Submerged arc welding (SAW) (8).

Other welding processes of some interest are gas welding with an oxy-acetylene torch used as a source of heat with a filler rod to deposit metal; electro-slag welding; and non-shielded arc.

The shielded metal arc uses a wide variety of coated electrodes that are classified by the strength level of the undiluted weld metal and type of coating. These different coatings vary the penetration of the electrode, welding positions and the amount of hydrogen in the gas shield. A detailed list of these electrode coatings is furnished subsequently. The SMAW process cannot be made automatic because the coating on the electrode does not permit these to be coiled and fed automatically. An adaptation of this process places the coating or flux constituents inside of a tubular electrode and permits automatic use but the variety of fluxes and resulting atomospheres available are much more limited. Both the GMAW and GTAW processes can be operated automatically or semi-automatically. The GMAW (or MIG) process is used on steels with helium, argon or mixed gas shielding and also with CO_2 shielding. The CO_2 gas is considerably cheaper than the others and is employed on steels when the mild oxidizing effect of this atmosphere can be tolerated. The GTAW (or TIG) process is slower than the other four methods listed and expensive because of the shielding gas. For these reasons, its use is restricted to cases where it has a technical advantage such as thin sections. These MIG and TIG processes have the advantages of being automatic with continuous feeding of the electrode into the weld and avoiding the fused flux or slag layer produced by the SMAW and submerged arc (SAW) processes. The submerged arc process is also automatic and is capable of substantially higher rates of metal. deposition and of making single pass welds of considerable thickness compared to the other three methods.

Gas welding is useful for small repair jobs and permits close control of the temperatures. Carbon steels can be welded without a flux but the rates of metal deposition is so slow that the process is not widely used in repairing or assembling steel castings, except for cladding and special purposes. The electroslag process is well adapted to joining thick sections (over 2 inches). It utilizes filler wires fed into a molten slag pool contained between water cooled dams. It starts as an arc but when the flux melts, the process depends on the heat generated by the flow of current from the electrode through the flux to the work. The process generates considerable heat, resulting in a coarse grained weld deposit and HAZ so that subsequent heat treatments are generally required. While not used extensively in casting repair up to this time, it is widely employed for the cast-weld assemblies with heavy sections (2). Non-shielded arc provides a weld of inferior quality that is only used for lower grade products.

Following the joining operation, a postweld heat treatment may be needed, depending on the base metal composition, welding method and service conditions. This treatment can range from stress relief of the weld area to quenching and tempering the entire weldment. Some of the purposes include: stress relief, improved strength and toughness, diffusion of hydrogen from the HAZ or restoring the corrosion resistance.

Types of Steels

For the purposes of this paper, the carbon steels will be considered in one category and alloy steels with a total alloy content of up to 5% in other groups. The alloy steels discussed contain intentional additions of the elements Ni, Cr, Si, MO and Mn. These added elements improve mechanical properties, heat treatment response, elevated temperature properties and corrosion resistance.

Plain carbon cast steels rely primarily on carbon for variation of their properties. The carbon content ranges between 0.10 and 1.00%, depending on the intended use. They can be welded without difficulty when the carbon is below about 0.30%,The weldability of plain carbon steels decreases as the carbon content is raised. The weldability of low-alloy steels is good, but they require careful attention to welding procedures and filler metal choice (5). As previously discussed, the strength, hardness and ductility of these steels varies widely depending on their composition and the thermal treatments. Both carbon and alloy steels are employed in the heat treated condition; the treatments used vary widely from annealing to normalizing to quenching and tempering, depending on the final use of the casting.

Low-Carbon Steels --

The low-carbon steels are considered to be those not exceeding 0.20% carbon. These are the most easily welded by a large variety of methods. Hardenability is very low and even though the maximum hardness for a fully-hardened structure may be considerable as illustrated in Figure 6 (4), the actual hardness in low-carbon HAZs rarely reaches high values because the critical cooling rates are not achieved under ordinary conditions. Preheating is not needed; several welding methods can be employed, and cooling is rarely encountered. However, the low-carbon steels are susceptible to porosity unless these are deoxidized. While it is usual practice to deoxidize low-carbon cast steels, many of the wrought steels are rimmed and this can lead to porosity from CO gas evolution unless electrodes with considerable amounts of aluminum, manganese and silicon are employed.

The low-carbon steels are welded commonly by the shielded metal arc, submerged arc and GMAW or MIG processes using CO_2 gas shield (9). Additional deoxidizers are usually added to the electrode in this latter case. The choice of electrodes for welding low-carbon steel is based on the desired final mechanical properties of the weld. Table I lists the AWS classification and weld deposit compositions of typical electrodes used in the SMAW process for carbon and low-alloy steels (4). The last two digits refer to the type of coating, while the preceding numbers list the minimum tensile strengths of the undiluted filler metal. Details on the characteristics of the various coatings are listed in Table II.

Electrodes of the EXXl0 and EXXll classes are suitable for low-carbon steels because of their deep penetration characteristic, minimal slag production. Low-hydrogen coatings are generally unnecessary

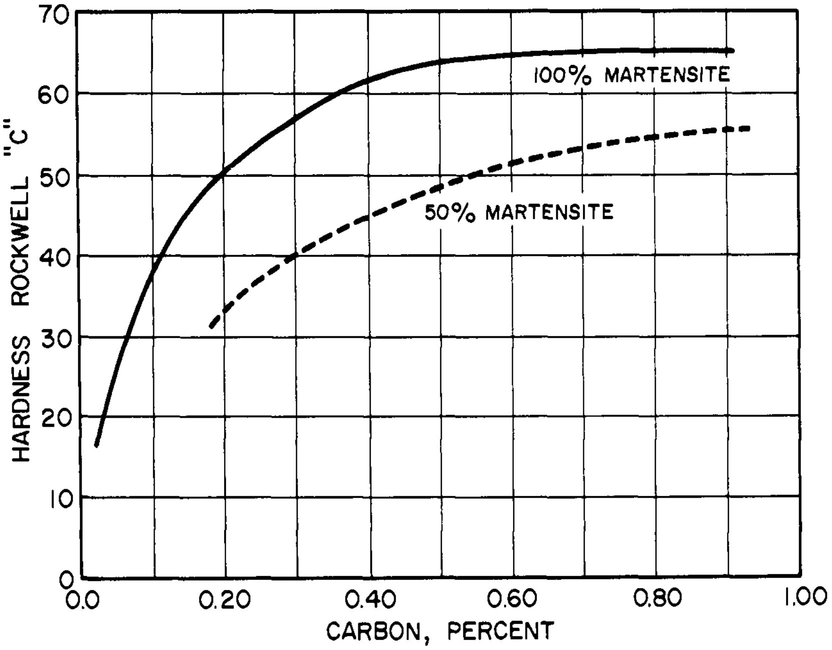


Figure 6. Relationship between carbon content and maximum hardness obtainable in any plain-carbon or alloy steel (4).

TABLE I - Chemical Composition Requirements and Typical Weld Deposit Analyses of Covered Electrodes for Carbon and Low Alloy Steels (4).

PART I — SEE "MILD STEEL COVERED ARC-WELDING ELECTRODES," AWS A5.1; ASTM A233

(A) THE FOLLOWING ELECTRODES HAVE NO COMPOSITION REQUIREMENTS IN AWS-ASTM SPECIFICATIONS

AWS-ASTM classification	*Typical Weld Deposit Analyses*							
	C	Mn	S	Si	Cr	Ni	Mo	V
E4510[a]	0.05	0.25	0.02	0.01	......	......	......	......
E4520[a]	0.05	0.25	0.02	0.01	......	......	......	......
E6010	0.06	0.45	0.02	0.20	......	......	......	......
E6010[a] iron powder	0.06	0.45	0.02	0.25	......	......	......	......
E6011	0.06	0.45	0.02	0.20	......	......	......	......
E6012	0.07	0.40	0.02	0.35	......	......	......	......
E6013	0.09	0.45	0.02	0.35	......	......	......	......
E6020	0.09	0.35	0.02	0.15	......	......	......	......
E6027	0.09	0.60	0.02	0.40	......	......	......	......
E6030[a]	0.09	0.35	0.02	0.15	......	......	......	......

(B) THE FOLLOWING ELECTRODES ARE REQUIRED BY AWS-ASTM SPECIFICATIONS NOT TO EXCEED THESE WELD DEPOSIT COMPOSITIONS:

	C	Mn	S	Si	Cr	Ni	Mo	V
E70XX series	......	1.25*	......	0.90	0.20*	0.30*	0.30*	0.08*

* The sum total of all elements with the asterisk shall not exceed 1.50 percent.

	Typical Weld Deposit Analyses							
	C	Mn	S	Si	Cr	Ni	Mo	V
E7014	0.08	0.90	0.02	0.25	0.10	0.05	0.03	0.01
E7015	0.08	0.90	0.02	0.50	0.10	0.05	0.03	0.01
E7016	0.08	0.90	0.02	0.50	0.10	0.05	0.03	0.01
E7018	0.08	0.90	0.02	0.60	0.10	0.05	0.03	0.01
E7024	0.08	0.90	0.02	0.60	0.10	0.05	0.03	0.01
E7028	0.08	0.90	0.02	0.60	0.10	0.05	0.03	0.01

[a] Not a recognized classification in current AWS-ASTM specifications.

PART II — SEE "LOW-ALLOY STEEL COVERED ARC-WELDING ELECTRODES," AWS A5.5; ASTM A316.

AWS-ASTM composition requirements[b] for weld deposits are listed opposite each classification. Single values shown are maximum percentages unless specific ranges are indicated. Typical weld deposit analyses when given are listed directly below the classification.

(A) CARBON-MOLYBDENUM STEEL ELECTRODES (E70XX)

AWS-ASTM classification	C	Mn	S	Si	Cr	Ni	Mo	V
E7010-A1	0.12	0.60	0.04	0.40	......	......	0.40–0.65	......
Typical	0.06	0.40	0.02	0.25	......	......	0.50	......

TABLE I - (Continued)

AWS-ASTM classification	C	Mn	S	Si	Cr	Ni	Mo	V
E7011-A1	0.12	0.60	0.04	0.40	......	......	0.40–0.65	......
Typical	0.06	0.40	0.02	0.30	......	......	0.50	......
E7015-A1	0.12	0.90	0.04	0.60	......	......	0.40–0.65	......
Typical	0.06	0.75	0.02	0.45	......	......	0.50	......
E7016-A1	0.12	0.90	0.04	0.60	......	......	0.40–0.65	......
Typical	0.06	0.80	0.02	0.40	......	......	0.50	......
E7018-A1	0.12	0.90	0.04	0.80	......	......	0.40–0.65	......
Typical	0.06	0.75	0.02	0.60	......	......	0.50	......
E7020-A1	0.12	0.60	0.04	0.40	......	......	0.40–0.65	......
Typical	0.06	0.45	0.02	0.25	......	......	0.50	......
E7027-A1	0.12	1.00	0.04	0.40	......	......	0.40–0.65	......
Typical	0.06	0.80	0.02	0.25	......	......	0.50	......

[b] Phosphorus content is limited to 0.03 max. percent in all cases, except where 0.030 max. percent is specifically indicated.

(B) CHROMIUM-MOLYBDENUM STEEL ELECTRODES (E80XX)

E8016-B1	0.12	0.90	0.04	0.60	0.40–0.65	......	0.40–0.65	......
Typical	0.10	0.75	0.02	0.40	0.50	......	0.50	......
E8018-B1	0.12	0.90	0.04	0.80	0.40–0.65	......	0.40–0.65	......
Typical	0.11	0.85	0.02	0.60	0.50	......	0.50	......
E8015-B2L	0.05	0.90	0.04	1.00	1.00–1.50	......	0.40–0.65	......
Typical	0.04	0.50	0.02	0.50	1.25	......	0.50	......
E8016-B2	0.12	0.90	0.04	0.60	1.00–1.50	......	0.40–0.65	......
Typical	0.06	0.50	0.02	0.50	1.25	......	0.55	......
E8018-B2	0.12	0.90	0.04	0.80	1.00–1.50	......	0.40–0.65	......
Typical	0.06	0.75	0.02	0.60	1.25	......	0.50	......
E8015-B4L	0.05	0.90	0.04	1.00	1.75–2.25	......	0.40–0.65	......
Typical	0.04	0.75	0.02	0.75	2.00	......	0.50	......

TABLE I - (Continued)

AWS-ASTM classification	C	Mn	S	Si	Cr	Ni	Mo	V
(C) NICKEL STEEL ELECTRODES (E80XX)								
E8016-C1	0.12	1.20	0.04	0.60	......	2.00–2.75	......	......
Typical	0.06	0.75	0.02	0.50	......	2.60	......	......
E8018-C1	0.12	1.20	0.04	0.80	......	2.00–2.75	......	......
Typical	0.06	0.75	0.02	0.60	......	2.60	......	......
E8016-C2	0.12	1.20	0.04	0.60	......	3.00–3.75	......	......
Typical	0.06	0.90	0.02	0.40	.....	3.50	......	......
E8018-C2	0.12	1.20	0.04	0.80	......	3.00–3.75	......	......
Typical	0.06	0.90	0.02	0.60	....	3.50	......	......
E8016-C3[c]	0.12	0.40–1.10	0.030	0.80	0.15	0.80–1.10	0.35	0.05
Typical	0.06	0.90	0.02	0.40	......	1.00	......	......
E8018-C3[c]	0.12	0.40–1.10	0.030	0.80	0.15	0.80–1.10	0.35	0.05
Typical	0.06	0.90	0.02	0.70	......	1.00	......	......
(D) CHROMIUM-MOLYBDENUM STEEL ELECTRODES (E90XX)								
E9015-B3	0.12	0.90	0.04	0.60	2.00–2.50	......	0.90–1.20	......
Typical	0.10	0.75	0.02	0.40	2.25	......	1.00	......
E9016-B3	0.12	0.90	0.04	0.60	2.00–2.50	......	0.90–1.20	......
Typical	0.10	0.75	0.02	0.50	2.25	......	1.00	......
E9018-B3	0.12	0.90	0.04	0.80	2.00–2.50	......	0.90–1.20	......
Typical	0.10	0.75	0.02	0.60	2.25	......	1.00	......
E9015-B3L	0.05	0.90	0.04	1.00	2.00–2.50	......	0.90–1.20	......
Typical	0.04	0.75	0.02	0.75	2.25	......	1.00	......
(E) MANGANESE-MOLYBDENUM STEEL ELECTRODES (E90XX, E100XX)								
E9015-D1	0.12	1.25–1.75	0.04	0.60	......	......	0.25–0.45	......
Typical	0.06	1.50	0.02	0.40	......	......	0.40	......

[c] Classification intended to conform to military specification for similar composition. Phosphorus limited to 0.030 max. percent.

TABLE I - (Continued)

AWS-ASTM classification	C	Mn	S	Si	Cr	Ni	Mo	V
E9016-D1[d]	0.12	1.25–1.75	0.04	0.60	......	......	0.25–0.45	......
Typical	0.06	1.50	0.02	0.40	......	......	0.40	......
E9018-D1	0.12	1.25–1.75	0.04	0.80	......	......	0.25–0.45	......
Typical	0.06	1.50	0.02	0.60	......	......	0.40	......
E10015-D2	0.15	1.65–2.00	0.04	0.60	......	......	0.25–0.45	......
Typical	0.12	1.85	0.02	0.40	......	......	0.40	......
E10016-D2	0.15	1.65–2.00	0.04	0.60	......	......	0.25–0.45	......
Typical	0.12	1.85	0.02	0.50	......	......	0.40	......
E10018-D2	0.15	1.65–2.00	0.04	0.80	......	......	0.25–0.45	......
Typical	0.12	1.85	0.02	0.60	......	......	0.40	......

[d] Not a recognized classification in current AWS-ASTM specifications.

(F) OTHER LOW-ALLOY STEEL ELECTRODES (EXXXX-G).

To meet the alloy requirements of this group, the weld deposit need have the minimum of one of the following elements:

EXXXX-G	......	1.00 min	......	0.80 min	0.30 min	0.50 min	0.20 min	0.10 min
				Typical Weld Deposit Analyses				
E8015-G	0.07	0.70	0.02	0.30	0.30	0.80	0.15	......
E8018-G	0.06	1.25	0.02	0.70	......	......	......	......
E9018-G	0.07	1.00	0.02	0.40	......	1.60	0.15	......
E9018-G	0.06	1.00	0.02	0.50	1.25	......	0.60	......
E10018-G	0.06	1.25	0.02	0.50	0.12	1.50	0.25	......
E10013-G	0.11	0.33	0.02	0.40	......	......	1.10	0.10
E10015-G	0.06	0.80	0.02	0.35	......	1.65	0.30	0.13
E11016-G	0.06	1.20	0.02	0.40	......	3.35	0.50	......
E11016-G	0.06	1.20	0.02	0.40	1.20	2.00	0.30	......
E11018-G	0.08	1.65	0.02	0.50	......	1.85	0.45	......
E11018-G	0.05	1.30	0.02	0.30	0.30	1.80	0.45	......
E11018-G	0.06	1.00	0.02	0.50	1.50	2.50	0.65	......
E11018-G	0.13	1.64	0.02	0.40	......	......	0.45	......
E12015-G	0.09	1.20	0.02	0.50	......	1.80	0.80	0.20
E12018-G	0.07	1.70	0.02	0.50	0.35	2.00	0.50	......
E12018-G	0.08	1.50	0.02	0.50	1.00	2.00	0.75	......

TABLE I - (Continued)

AWS-ASTM classification	C	Mn	S	Si	Cr	Ni	Mo	V
(G) CLASSIFICATIONS OF AWS A5.5 (ASTM A316) INTENDED TO CONFORM TO MILITARY SPECIFICATIONS FOR SIMILAR COMPOSITIONS.[e]								
MIL-7018[f]	0.12	0.40–1.25	0.030	0.80	0.15	0.25	0.35	0.05
MIL-8018[f]	0.12	0.40–1.10	0.030	0.80	0.15	0.80–1.10	0.35	0.05
E9018-M	0.10	0.60–1.25	0.030	0.80	0.15	1.40–1.80	0.35	0.05
E10018-M	0.10	0.75–1.70	0.030	0.60	0.35	1.40–2.10	0.25–0.50	0.05
E11018-M	0.10	1.30–1.80	0.030	0.60	0.40	1.25–2.50	0.30–0.55	0.05
E12018-M	0.10	1.30–2.25	0.030	0.60	0.30–1.50	1.75–2.25	0.30–0.55	0.05

[e] Phosphorus content limited to 0.030 max. percent for these electrodes.

[f] Not presently an AWS-ASTM classified electrode. See military specification MIL-E-22200/1B.

(H) LOW-ALLOY STEEL COVERED ELECTRODES FOR HEAT TREATABLE WELD DEPOSITS.[g]

Cr-Mo	0.10	0.55	0.02	0.45	0.50	......	1.10	......
Cr-Ni-Mo	0.20	1.50	0.02	0.50	0.50	1.25	0.25	......
4130	0.25	1.00	0.02	0.50	1.00	......	0.25	......
4140	0.40	1.00	0.02	0.50	1.00	......	0.25	......
4340	0.40	1.00	0.02	0.50	1.00	2.00	0.25	......
MIL-13018	0.10–0.15	0.80–1.15	0.030 max	0.30–0.60	0.90–1.20	1.50–2.00	0.45–0.75	0.02 max

[g] Not recognized classifications in current AWS-ASTM specification.

TABLE II - Shielded Arc Welding Electrode Coatings and Characteristics (4).

Electrode	Type of Covering	Penetration	Surface Appearance	Slag
EXX10	High-cellulose sodium	Deep	Flat, wavy	Thin
EXX11	High-cellulose potassium	Deep	Flat, wavy	Thin
E6012	High-titania sodium	Medium	Convex, rippled	Heavy
EXX13	High-titania potassium	Shallow	Flat or concave, slight ripple	Medium
EXX14	Iron powder, titania	Medium	Smooth, fine ripples	Medium
EXX15	Low-hydrogen sodium	Medium	Flat, wavy	Medium
EXX16	Low-hydrogen potassium	Medium	Flat, wavy	Medium
EXX18	Iron powder, low hydrogen	Medium	Convex, smooth even ripple	Medium
EXX20	High iron oxide	Medium	Flat or concave, smooth	Heavy
EXX24	Iron powder, titania	Shallow	Convex, smooth fine ripple	Medium
EXX27	Iron powder, iron oxide	Medium	Slight concave, smooth, even ripple	Heavy
EXX28	Iron powder, low hydrogen	Shallow	Flat to concave, smooth fine ripple	Heavy

unless sufficient residual alloy elements are present to raise the composition to low alloy levels. The low-hydrogen coatings require careful handling and may pick up moisture when exposed to air for any significant length of time. If these become moist, they should be oven dried and taken hot from the oven for usage. Therefore, the electrodes most often used are the cellulosic electrodes at strength levels of 60 ksi. These electrodes give a good combination of mechanical properties and ease of welding at all positions (9).

Most problems with low carbon steels are caused by impurity elements and non-homogeneous structures. Sulfur is the most common and damaging impurity element. Sulfur levels at 0.05% can produce liquid sulfide films at the grain boundaries (10) that cause solid embrittling films and possible hot cracking, although this is a much more serious problem in the higher carbon and alloy grades. The presence of strong sulfide formers, such as manganese, reduce this problem but may not completely eliminate it (ll). The presence of manganese sulfide inclusions also reduces problems with hydrogen embrittlement by providing sinks for this element (12, 13). Other more stable sulfide formers, such as rare earths, can be more effective in this respect. Phosphorus has a similar embrittling effect to sulfur (4, 14) and is more difficult to remove.

These problems with hot cracking caused by low melting point films and embrittling elements apply to the entire range of carbon and alloy steels. In fact, sulfur and phosphorus levels are much more critical in higher performance steels, and maximum allowable contents are generally much lower than in low-carbon steels. The higher tolerance of low-carbon steels for impurity elements exists because the segregation of sulfur and phosphorus are increased by higher carbon contents and certain alloy elements. The only low-carbon steels which require some precautions are the free-machining steels, where sulfur levels may reach 0.30%. Even though additional manganese is specified in these steels, welding with cellulosic electrodes or other conditions producing hydrogen may result in porosity, due to the formation of H_2S gas in the weld metal. Electrodes of the EXX15, EXX16 and EXX18 classes may be used to avoid this problem.

Low-carbon steels are not highly susceptible to cold cracking even under fairly severe conditions; the lack of toughness characteristic of higher-carbon martensites is not found in these cast steels under 0.20% C.

Medium Carbon Steels --

Medium carbon steels are generally considered to be plain carbon steels with 0.20-0.50% carbon.Hardening of the HAZ so that preheating is required is generally not necessary below 0.30% carbon. The suggested composition limits above which preheating is required are 0.28% carbon and 1% Mn according to one source (14). As the carbon increases above 0.30%, preheating and the use of low-hydrogen electrodes become more of a necessity to avoid cracking problems. Towards the top of this range, the HAZ often attains full hardness without additional alloying and reaches the maximum hardness shown in Figure 6. This high hardness with a low ductility or brittleness make the hardened structures very susceptible to both cold-cracking from thermal stresses and to hydrogen-induced cracking. Figure 7 (4) shows underbead crack sensitivity, underbead hardness and bend angle at maximum load as a function of carbon equivalent. This carbon equivalent has been formulated to allow for the influence of the manganese and silicon contents on the hardening behavior of these medium carbon steels. It is evident that both cracking susceptibility and hardness increase rapidly with carbon equivalent and that the bend angle at failure decreases. All these trends denote a marked loss in toughness.

Close control of cooling rates in the HAZ must be maintained. The preheat temperature required depends on the carbon content, section thickness and number of weld passes. These temperatures vary from 100°F for thin sections at 0.35% C to 400°F for 0.45% C with 4 inch sections and single pass welds. The preheat requirements for multipass welds are substantially less if the welding operations are conducted efficiently. Preheating temperatures of 400°F are needed with 0.50% C steels welded under other than low-hydrogen conditions and temperatures of 500°F are used in the upper part of the 0.30-0.50% C range for heavy sections, single pass welds and other than low-hydrogen conditions (1,4). Interpass temperatures at least equaling the preheat temperature should be maintained.

High Carbon Steels --

High carbon steels are classified as containing between 0.50 and 1.00% carbon. They are difficult to weld because of cracking in the weld area from the high hardness and brittleness that can result from martensite formation. Precise determination of the necessary preheat and interpass temperatures is difficult for steels with carbon contents over about 0.50% C because of their sensitivity to other factors. Attempts have been made to correlate hardness and crack sensitivity directly, but the results indicate that such a relationship cannot be isolated (4). The following welding procedure is recommended: low-hydrogen shielding when available in desired strength levels, a welding technique which minimizes dilution and the resultant hardening of the weld metal and a postweld heat treatment consisting of at least a stress relief and possibly a full anneal (1). The choice of a filler metal depends on the desired final strength of the joint. Weldments used in the quenched-and-tempered condition require E90XX or El00XX electrodes; lower strength filler metals may be used in other cases.

Preheating should always be used, the exact temperature being determined by cooling conditions and welding method. Temperatures of 600°F or even higher are sometimes employed, but such temperatures cause difficulties with oxidation and welder discomfort. The preheat temperature must be adjusted in such cases to allow the diffusible hydrogen to leave the hardened area and prevent additional embrittlement. In addition to the preheat, welding techniques that provide slow cooling should be used including a high heat input, multipass welds and the use of insulation to restrict heat flow.

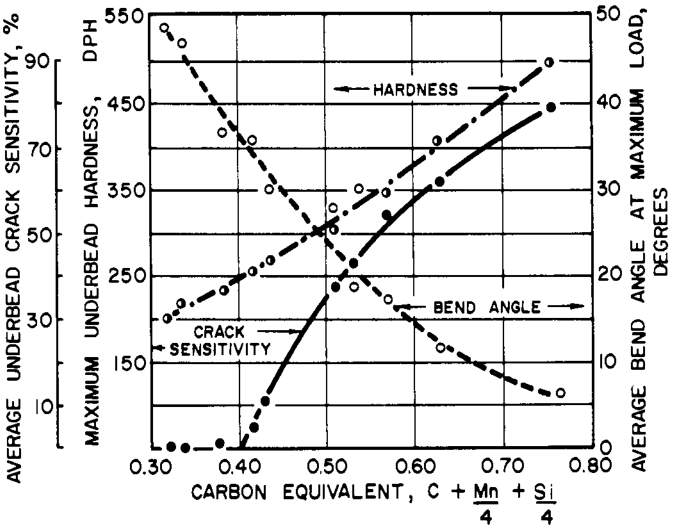


Figure 7. Effect of carbon equivalent on underbead hardness, crack sensitivity and bend angle at maximum load (4).

Alloy Steels --

Classification of the weldability of alloy steels by composition is difficult since many different combinations and purposes of alloying are employed. Alloys are added to steels to improve the strength, toughness and corrosion resistance. The influence of alloying on weldability has been equated to carbon content by a term known as carbon equivalent. This approach was used for the other elements in carbon steel in Figure 7. A fairly widely used relation is shown below(8).

$$\text{Welding Carbon Equivalent} = \%C + \frac{\%Mn}{20} + \frac{\%Ni}{15} + \frac{(\%Cr + \%Mo + \%V)}{10}$$

As this value increases, the susceptibllity of the steel to cold-cracking does likewise. The carbon equivalent is designed to be employed along with a knowledge of the electrode size and weld dimensions as a means of selecting preheating temperatures. While this approach has the advantage of simplicity, it is limited in range and is an oversimplification of the situation.

Even within the limits of 5% total alloy content employed in this paper, a considerable variety of alloy steel castings are included. These have a wide range of weldability and are utilized for many purposes. Considerable detail on their repair welding is presented in reference 1, and it is not the purpose of this publication to duplicate that coverage. A summary of the electrodes used, preheat and postheating temperatures recommended are included in Table III (15). This table provides limited data on the carbon steels already discussed, as well as the various categories of low and medium alloy steels.

The majority of the alloying elements employed increase the hardenability of the steel and reduce the critical cooling rate. In the more highly-alloyed steels, full hardening is possible even in thick sections and at modest cooling rates. With this increase in hardenability, the carbon content becomes increasingly critical because it controls the hardness of the martensite and affects its toughness. The relation between maximum hardness and carbon content, as shown in Figure 6, becomes a realistic estimate of the hardness of martensitic alloy steels because fully-hardened structures can readily be obtained. A carbon content of 0.20% in a plain carbon steel might produce a hardness of R_c 30 under relatively rapid cooling conditions, whereas the same carbon level in a high-alloy steel at the same cooling rate might result in a hardness of R_c 40-45. Such hardness levels markedly increase the susceptibility to cold-cracking and generally require a tempering treatment before use in service.

In addition to increasing hardenability, most alloy elements except for silicon, cobalt and aluminum, lower the M_s temperature of the steel. Silicon has no effect and aluminum and cobalt raise the M_s. An expression for calculating the M_s temperature for a steel of known composition as shown below (4):

$$M_s(°F) = 1042 - 853(\%C) - 60(\%Mn) - 30(\%Ni) - 30(\%Cr) - 38(\%Mo)$$

TABLE III - WELDING PROCEDURES FOR CARBON AND ALLOY CAST STEELS(15)

Type of Cast Steel	Electrodes		Preheat Temperature in Degrees F	Postheat Temperature in Degrees F	COMMENTS
	Classification	Diameter** in Inches			
Carbon 0.15-0.24%	All E60XX	3/16	None	No requirements	
Carbon 0.25-0.33%	E70XX; E80XX for higher carbons	3/16	None except for widely varying sections	No immediate care needed	
Carbon 0.34-0.50%	E70XX, E80XX or E90XX; for Q & T omit E70XX	3/16	None if EXX15 or EXX16 used, otherwise use 200-400	None except for over 0.40% C. All 11-1200 stress relieved***	SAE prohibits welding of carbon over 0.35% Higher C castings should be cooled slowly
Nickel 2300 2-3%; 3-3.75%	E70XX to E100XX	3/16	None with EXX16 to 0.35% C; other electrode: 200-400	Cool in still air; stress relieve 1100-1250	High nickel and/or carbon - stress relieve immediately after welding or furnace cool
Ni-Cr (3100)	E90XX or 100XX E120XX for high C and high alloy	3/16	None for low C with EXX16 200-400	Low C & alloy - still air; most - cool in furnace or under insulation	Low hydrogen electrodes permit lower preheat; may stress relieve immediately after welding, 1100-1250
Ni-Cr-Mo 4300	Use low hydrogen E100XX-120XX	3/16	400-600	Insulate or furnace cool, or stress relieve after welding	Full heat treatment after welding, low carbon - stress relieve 1100-1250
Ni-Cr-Mo 8600 9800	Low hydrogen E90XX; E100XX; high C - E120XX	3/16	Low carbon EXX16-none; High C 200-400	Cool in still air, but retard cooling furnace, cool if not stress relieved	Castings should be ultimately stress relieved 1100-1250, higher carbon steels full heat treatment
Medium Mn 1-2%	E90XX & E100XX E120XX for high C	3/16	None for low C and low alloy	Cool in still air high C - furnace cool or insulate	Low hydrogen electrodes eliminate preheat in many cases; all castings should be stress relieved 1100-1250
Mn-Mo Mn 1.00-1.35% Mo 0.10-0.30%	E90XX & E100XX E120XX for High C	3/16	200-400	1100-1200	Use same welding precautions as for medium manganese steel castings
Mn-Cr-Mo Mn 1.25-1.50% Cr 0.70-0.90% Mo 0.30-0.40%	E80XX to E100XX E120XX for some high C - higher alloy	3/16	Low alloy - none with EXX16; others 200-400	Cool in still air high C & alloy furnace cool or immediately stress relieve 1100-1200	Castings should be ultimately stress relieved. Full heat treatment, if possible
Cr-Steel 5100	E90XX, E100XX & occasionally E120XX use low hydrogen if possible	3/16	Low C & low alloy, none if use EXX16; others 200-400	Still air, high C & alloy furnace cool or with insulation	Castings should be stress relieved 1100-1250
Mo-Steel 4000	E80XX to E100XX	3/16	Low C with EXX16 none Higher C 200-400	Low C still air; others - F.C. or use insulation unless stress relieved after welding.	For heavy sectioned castings, 400 to 600 degrees F Preheat. All castings stress relieve 1100-1250 degrees F.

*Fuller details are given in "Recommended Practices for Repair Welding and Fabrication Welding of Steel Castings".[1]

**Maximum diameter for vertical and overhead welding. Otherwise use largest diameter which can be readily handled.

***The stress relieving treatment can be used only if it does not reduce the strength or hardness of the casting below its requirements.

The hardenability is affected by grain size and segregation as well as composition and requires special calculations that are not presented in this paper.

Alloying elements increase the susceptibility of steels to welding discontinuities, particularly cracking and the metallurgical types. The susceptibility to hot cracking is still largely controlled by the sulfur content of the steel but alloy steels are more sensitive to a given sulfur level. While a 0.05% S level is permissible in many carbon steels, a much lower sulfur content can cause serious cracking problems in an alloy steel. The effects of alloys on hot cracking susceptibility depend primarily on their influence on the behavior of sulfur. Those that promote sulfide segregation to grain boundaries by reducing the solubility of sulfur also promote hot-cracking. Such elements include carbon, phosphorus, nickel, aluminum and large amounts of silicon (over 1% and only in the heat-treatable alloys). As mentioned above, manganese tends to form fairly stable sulfides which resist grain boundary segregation and therefore decrease hot-cracking susceptibility. Other elements which reduce hot-cracking tendencies are chromium, vanadium and molybdenum. These elements exert their influence indirectly by forming stable carbides and thus decreasing the effect of carbon. While the general influence of all these elements is known individually, it has not been possible to predict their combined effects. For this reason, the cracking tendency of each particular steel is evaluated by means of standard weldability tests.

The presence of significant amounts of alloys increases the susceptibility of the steels to both cold-cracking and hydrogen-induced cracking. The higher hardenability in alloy steels results in structures which show less toughness in the as-welded condition than carbon steels of similar carbon content. The increased cracking results from a number of factors: higher yield strengths obtained by alloying result in increased stored elastic energy; the lower M_s and M_f temperatures of alloy steels mean that transformation often takes place at temperatures which cannot act to temper the martensite, stress relieve, or produce significant diffusion of hydrogen from the hardened area; and the more complete transformation to martensite, which has a lower solubility for hydrogen than austenite or ferrite, results in a greater degree of supersaturation of hydrogen. It is evident that great care must be observed in welding alloy steels to prevent cold-cracking and only low-hydrogen processes be used when possible.

The choice of filler metal for alloy steels is more critical in alloy than carbon steels. Whereas low or medium carbon steels could be welded with a number of filler materials, alloy steels frequently have properties that can only be obtained by a limited number of weld metals. The selection is further complicated by the fact that matching of compositions is often impossible or undesirable. The filler metal requirements must be examined from the standpoint of weld metal properties in the final structure rather than composition and properties of filler metal alone. Increased alloy content of filler metal produces a problem with oxidation losses and the weld metal composition is greatly affected by pickup of alloy elements from the base metal.

In addition to the general coverage provided by Table III and the above discussions, some individual considerations are the heat treated condition in which the casting is received and the subsequent heat treatment that is to be conducted. In cases where field weldability without preheating or postheating is desired, the alloy steels have a low carbon content (generally below 0.20%) and depend on multiple Mn-Cr-Mo-Ni Alloys for their strength. The purpose is to maintain a low carbon equivalent value. The low carbon level provides a much tougher, crack-resistant martensite when this is formed. Depending on the strength level and the section thickness, the amount of alloying and heat treatment employed prior to welding is adjusted. These steels are welded in the normalized, normalized and tempered, and quenched and tempered conditions. When postheating procedures are not feasible, it is the recommended welding practice to weld with low hydrogen electrodes and to provide fast cooling rates in the HAZ during welding. This technique deposits small weld beads at low heat input in a stringer fashion with no weaving of the electrode. The multiple weld deposits provide a self-tempering effect. The strength levels of these field welded steels are, of course, limited because the amount of hardening elements is restricted by the requirements for a relatively low carbon equivalent.

When the application requires higher strength or involves heavier sections, the castings are quenched and tempered after welding. However, in cases where the quenched and tempered weld metal fails to reach the required strength, only tempering can be used. These castings are frequently annealed prior to welding and require preheating temperatures as high as 600°F. Since these steels are to be heat treated after welding, high heat inputs during welding are used to retard martensite formation. The deposited weld also is usually high strength in these cases to provide the needed properties in the finished component. Cracking of this deposited weld then becomes a more severe problem than the HAZ of the base casting making the use of the higher preheating temperatures a necessity (8).

III. MECHANICAL PROPERTIES OF WELDS

In addition to the strength and ductility requirements, the dynamic properties of repair and fabricated welds in steel castings are of major significance in the service behavior of the finished components. Because welds present an altered dimensional and metallurgical structure, as discussed in the previous section, the properties of welds are frequently considerably different than those of the base, heat treated casting. Many of these effects are produced by the metallurgical discontinuities that have been described since the welds have both a different structure and composition compared to the base metal. The presence of the other types of weld discontinuities (Figure 1) also exert considerable influence on mechanical behavior. This part in the paper discusses the effect of numerous welding processes, variables and discontinuities on the fatigue and toughness properties of welds in steel castings.

Fatigue Behavior

Weld Configuration --

The weld geometry or configuration is a major consideration in weldments joining wrought sections and shapes but these considerations do not apply for the most part to welds in steel castings. When repair or fabrication welds are made on steel castings and ground flush to the smooth contour of the part, the effect of weld geometry is removed unless undercuts have occurred during welding. However, when fillet welds are employed to make "L", "T", "X" or box sections, and the welds are not ground smooth, the weld geometry can influence fatigue strength markedly. A report has been published by SFSA showing the superiority of the fatigue properties of cast "L" and box sections compared to those shapes on welded joints in medium carbon steels (16). The problem of the fatigue behavior of weld configurations in fatigue and the desirability of using steel castings or forgings for some contours has been recognized in other texts (17).

The effect of not grinding the weld reinforcement or raised weld deposit from a butt weld is illustrated in Figure 8 (17). This figure also indicates that if the weld reinforcement is removed completely by proper grinding or machining, the fatigue strength of the joint starts to approach the properties for the as-received base plate. Fatigue tests conducted on simulated butt-welded specimens of mild carbon structural wrought steel with the simulated weld reinforcement machined from base metal showed similar results (18). The significant variables are the flank angle, 8, and the radius at the toe of the weld, R as shown in Figure 9a. Decreasing the flank angle or increasing the radius will increase fatigue strength as illustrated in Figure 9b (18). Other weld configurations, such as fillet welds, are also influenced by the shape of the weld reinforcement, the overall dimensions of the weld and the presence of small undercuts at the toe of the weld (16,17). The fatigue properties of welds and lap joints are even more affected by geometrical considerations but these shapes usually do not apply to steel castings.

Base Metal Strength Level --

The unnotched fatigue strength of carbon, low and medium alloy steel castings increases with tensile strength, at least up to a tensile strength of about 200 ksi. The notched fatigue strength, however, begins to level off at a lower strength (160-180 ksi U.T.S.). This behavior is illustrated in Figure 10 (15,19). The notched fatigue test results show more scatter than the unnotched fatigue strengths and are usually considered to be more applicable to the service performance of most components.

Welding may reduce the fatigue strength of the higher strength steels even more than the presence of mechanical notches. Data in the welding literature (20-23) show that:

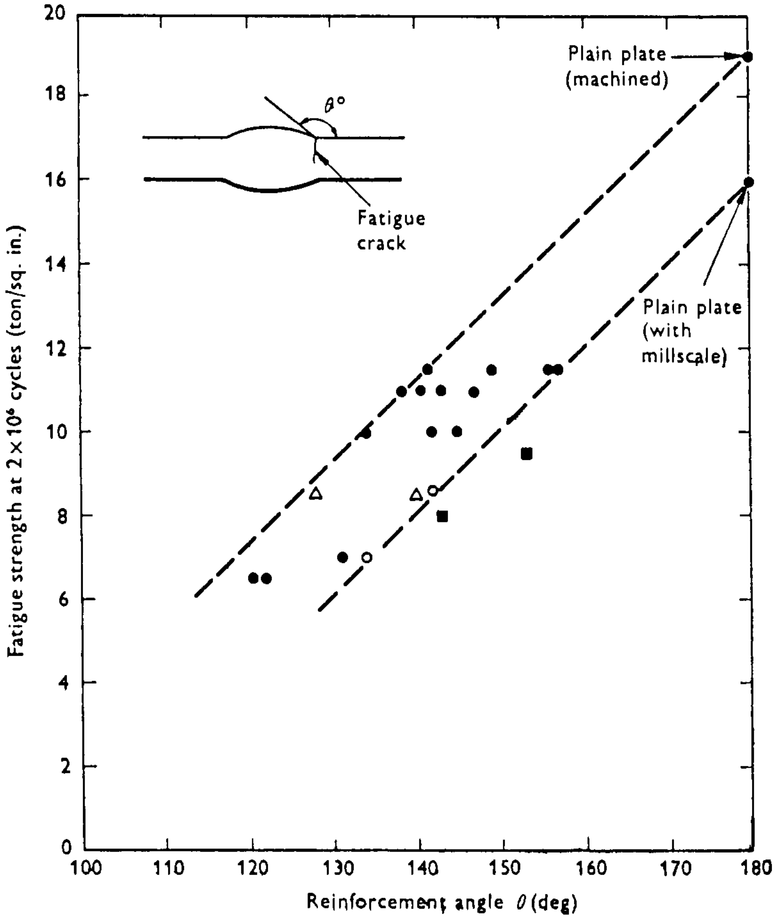


Figure 8. The relationship between reinforcement angle and fatigue strength of transverse butt welds (17).

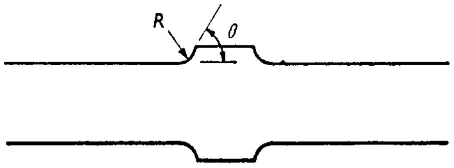


Figure 9a. Type of specimen used showing dimensions plotted in graph 9b.

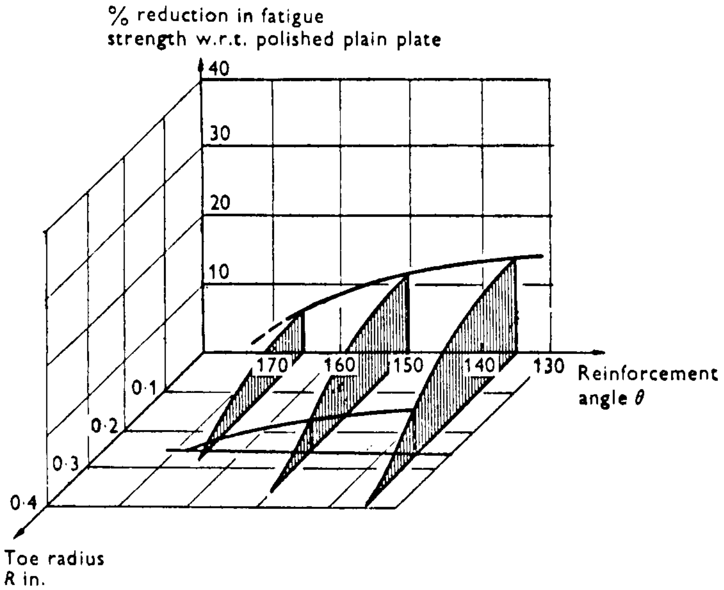


Figure 9b. The effect of reinforcement shape in tests subjected to alternating loading (18).

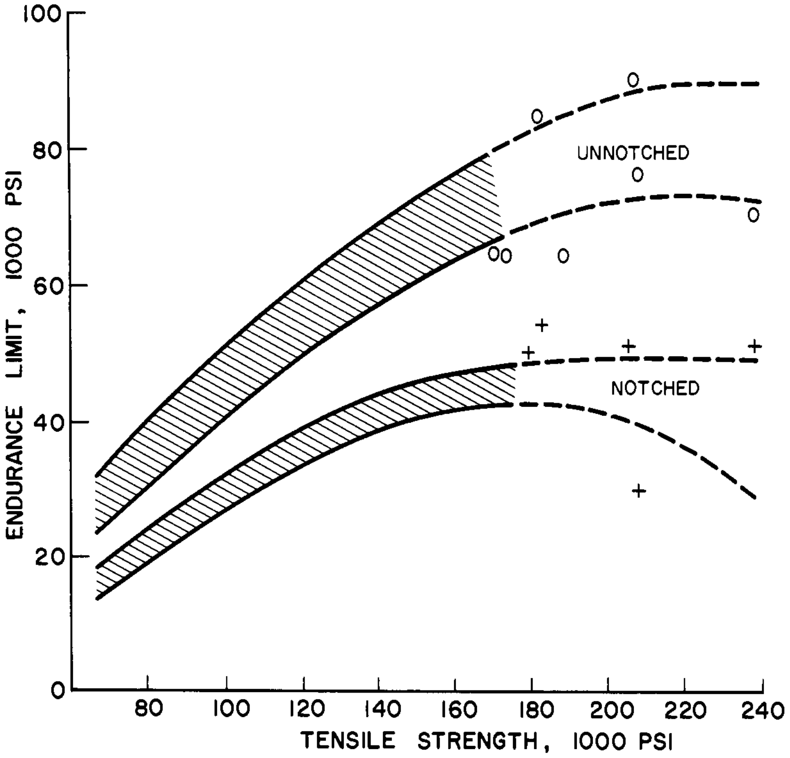


Figure 10. Endurance limit for notched and unnotched fatigue specimens of cast steel of various tensile strength (15, 19).

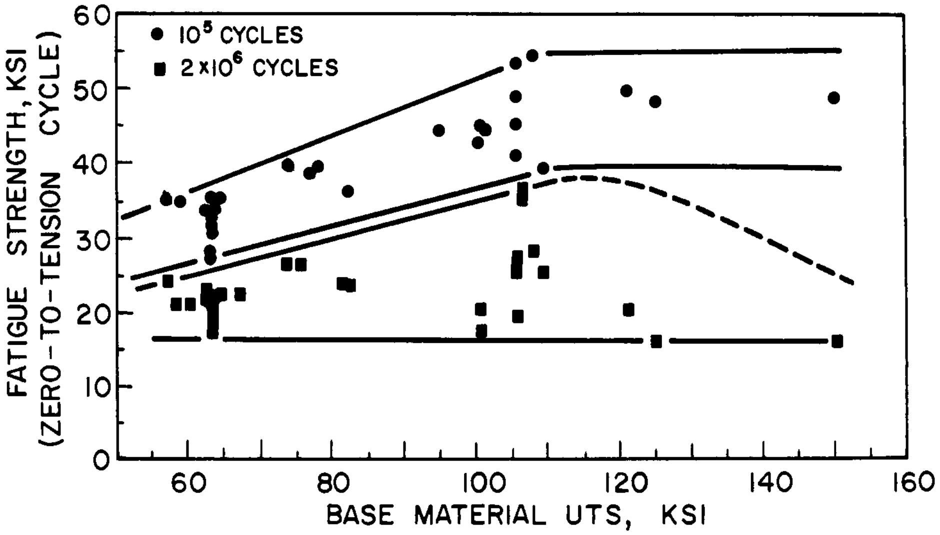


Figure 11. Effect of base metal tensile strength on weld fatigue strength for transverse butt welds tested in pulsating tension (R=0) (20).

(1) for steels with weld strengths of 55 to 110 ksi, weld fatigue strength increases slightly with increasing tensile strength;

(2) the fatigue strength of higher strength steels levels off because of the increasing notch sensitivity at higher tensile strengths;

(3) considerable scatter occurs in the data resulting in reports that the fatigue strength is not affected by tensile strength.

This effect of the tensile strength of the base material welded on the fatigue strength of transverse butt steel welds is illustrated in Figure 11 (20). In this and in subsequent plots the R value refers to the algebraic ratio of the smallest and largest stress (tension +, compression -) during fatigue testing. In this case, the testing cycle was zero stress-tension.

An explanation has been presented for the fact that the fatigue strength of weldments does not show the same degree of dependence on tensile strength exhibited by the notched fatigue strength of the base metal (24). It was reported that fatigue cracks which form in the region of stress concentration at the toe of the weld actually initiate at non-metallic inclusions which further concentrate the stress. This high stress concentration makes the stress at the inclusion tip very high and the strain independent of material strength. Since fatigue is strain dependent, it too becomes independent of material strength.

High and Low Hydrogen Electrodes --

The earlier investigators of this subject concluded that the use of low-hydrogen electrodes resulted in no advantage in fatigue strength over that for high hydrogen rutile electrodes. In one case (25), it was stated that no significant difference exists between the strength of butt welds made with rutile electrodes and welds made with low-hydrogen electrodes provided that they both have the same reinforcement shape.

However, when the weld reinforcement is removed, the advantages of low hydrogen welds become evident. With the reinforcement intact, the fatigue crack initiates at the toe of the weld in the HAZ of the base plate. When the reinforcement is removed, fatigue cracks initiate at discontinuities in the weld and the properties of the weld material become significant. Fatigue tests made on all weld metal fatigue specimens machined from welds indicate the superiority of low hydrogen weld metal (26). In additional tests, the weld overfill or reinforcement was machined off, and a hole was drilled through the weld to produce a stress concentration within the weld metal (27). When these specimens were tested in pulsating tension, welds made with low hydrogen electrodes displayed significantly higher fatigue strengths. Similar results are also available for machined butt welds with notches 0.79 inches deep and a 0.10 inch radius machined in the edge of a

3 inch wide, 24 inch long specimen as illustrated in Figure 12 (26). Other data on the superiority of welds made with low hydrogen electrodes in the presence of weld discontinuities will be discussed in the following section.

Weld Discontinuities --

An extensive investigation (28) has been undertaken by the SFSA to determine the influence of welding discontinuities on the fatigue strength of welded cast steel in bending and torsion. The tests were all performed on low alloy (8630) steel that was normalized and tempered to a tensile strength level of about 88 ksi and quenched and tempered to approximately 122 psi. The dimensions of the fatigue specimens employed are shown in Figure 13 (28).

These specimens were welded in the as-cast condition, and the normalizing and tempering and water quench and tempering treatments were performed (austenitizing in salt pots) after welding was completed. Accordingly, these results show the effects of the weld discontinuities only, and are not influenced by the heat effect that occurs during welding. Double vee butt welds were employed on each type of specimen using the shielded metal arc welding process and commercial E9018-B3, low hydrogen type electrodes. A preheat temperature of 300°F was used to produce slow cooling and post-heat of ll00°F for one hour immediately after welding to prevent cold cracking.

The weld joint, number and order of weld bead deposition and the types of discontinuities tested are shown in Figure 14 (28). The classification of the welding discontinuities that were tested are listed in Table IV (28). The results of the fatigue tests are summarized for bending in Table Va and for torsion in Table Vb. In each case, the test results are compared with a sound cast steel of similar heat treatment and strength level.

This investigation (28) conducted complete S-N curves on these steels, so the values of K_f* were obtained for the various types of discontinuities on the normalized and tempered and quenched and tempered steels for both bending and torsion fatigue tests. These K_f values have been listed in Table VIa for bending and Table VIb (28) for torsion at 10^5, 10^6, and 10^7 cycles,

The test results indicate that undercuts have the most marked effects on fatigue life, followed by slag inclusions and incomplete penetration with the least effect from the sound and sound weld tests machined flush to the fatigue bar surface. The figures below indicate the approximate percentage loss in endurance limit from the various discontinuities for bending and torsion fatigue (28, 28a).

$$*(K_f) = \frac{\text{Endurance Limit of Unnotched Specimens}}{\text{Endurance Limit of Notched Specimens}}$$

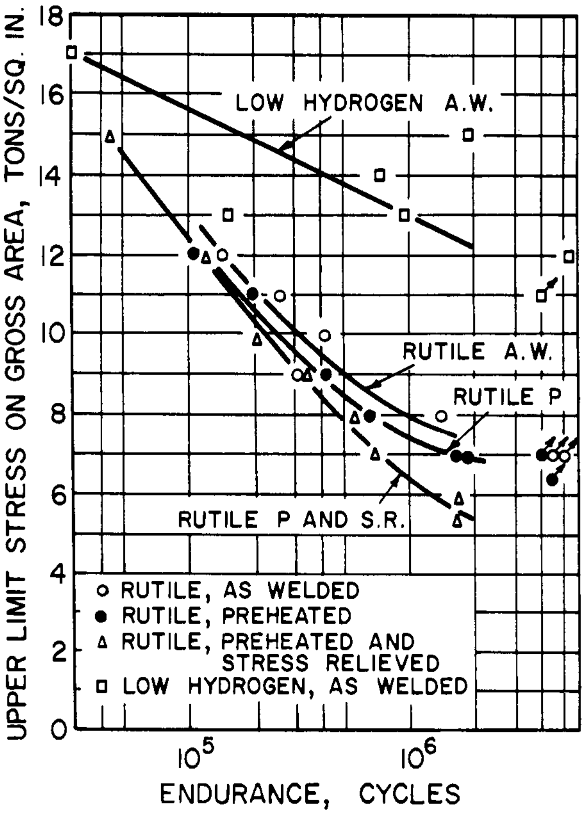
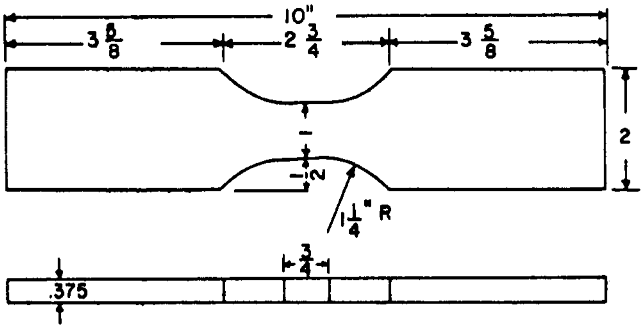
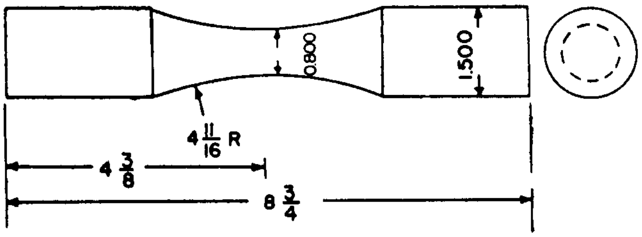


Figure 12. Fatigue test results for machined butt welds with V edge notches (26).

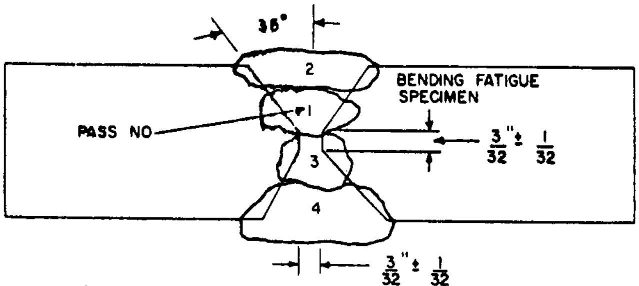


a) PLATE BENDING FATIGUE SPECIMEN

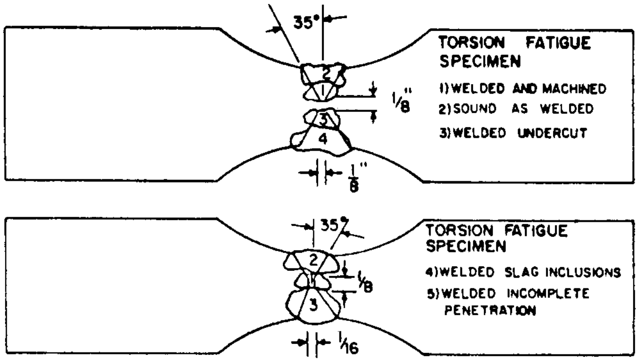


b) TORSION FATIGUE SPECIMEN

Figure 13. Dimensions of the bending and torsion fatigue specimens (28).



a. Bending Fatigue Specimen.



b. Torsion Fatigue Specimen.

Figure 14. Sketch of weld joint design (not to scale) for fatigue specimens (28).

TABLE IV: ASTM CLASSIFICATION OF DISCONTINUITIES IN FATIGUE SPECIMENS (28).

Discontinuity	Severity (ASTM Classification)	Unacceptable for Casting or Weld Classes
Weld - Incomplete Penetration	5-D (E99-55T) VI-2 (E125-56T)	1 through 4
Weld - Slag	3-D (E99-55T) VI-4 (E125-56T)	1 through 4
Weld - Undercut	9-E (E99-55T) VI-3 (E125-56T)	1 through 5

TABLE Va: SUMMARY OF FATIGUE RESULTS FOR BENDING FATIGUE SPECIMENS(28)

Heat Treatment	Tensile Strength (psi)	Discontinuity	Endurance Limit (psi)	Endurance Ratio
QT	145,000	Cast Steel - Sound	45,000	0.310
QT	122,000	Weld-Machined - Sound	30,600	0.251
QT	122,000	Weld - Slag	29,700	0.243
QT	122,000	As Welded - Sound	29,450	0.241
QT	121,200	Weld - Incomplete Penetration	29,100	0.240
QT	121,200	Weld - Undercut	28,300	0.233
NT	83,100	Cast Steel - Sound	30,000	0.361
NT	89,800	Weld - Machined - Sound	31,500	0.352
NT	88,300	Weld - Incomplete Penetration	31,100	0.350
NT	89,800	As Welded - Sound	30,900	0.345
NT	89,900	Weld - Slag	28,200	0.314

QT = Quenched and Tempered

NT = Normalized and Tempered.

TABLE Vb: SUMMARY OF FATIGUE RESULTS FOR TORSION FATIGUE SPECIMENS(28)

Heat Treatment	Tensile Strength (psi)	Discontinuity	Endurance Limit (psi)	Endurance Ratio *
QT	126,000	Cast Steel - Sound	27,600	0.298
QT	124,100	Weld - Machined	28,500	0.230
QT	124,100	As Welded - Sound	27,400	0.221
QT	121,200	Weld - Undercut	23,600	0.195
QT	124,100	Weld - Slag Inclusions	22,700	0.184
NT	83,100	Cast Steel - Sound	23,600	0.270
NT	90,200	Weld - Machined	23,600	0.261
NT	90,200	As Welded - Sound	22,600	0.250
NT	90,200	Weld - Slag	21,200	0.234
NT	88,300	Weld - Undercut	20,300	0.230

* $\frac{\text{Endurance Limit}}{\text{Tensile Strength}}$ = Endurance Ratio

TABLE VIa: K_f VALUES FOR SEVERE DISCONTINUITIES IN BENDING FATIGUE AT VARIOUS VALUES OF ENDURANCE LIFE(28).

Type of Specimen	Endurance of Life (Cycles) 10^5	10^6	10^7
QT			
Weld - Machined	1.33	1.14	1.13
As Welded - Sound	1.52	1.23	1.17
Weld - Incomplete Penetration	1.49	1.23	1.18
Weld - Slag	1.35	1.19	1.16
Weld - Undercut	1.46	1.23	1.21
NT			
Weld - Machined - Sound		1.04	1.03
Weld - Incomplete Penetration	1.23	1.03	1.03
As Welded - Sound	1.32	1.12	1.05
Weld - Slag	1.38	1.20	1.15
Weld - Undercut		1.29	1.29

TABLE VIb: K_f VALUES FOR TORSION FATIGUE SPECIMENS AT VARIOUS VALUES OF ENDURANCE LIFE(28).

Type of Specimen	Endurance of Life (Cycles) 10^5	10^6	10^7
QT			
Weld - Machined - Sound		1.22	1.30
As Welded - Sound	1.37	1.35	1.35
Weld - Undercut	1.50	1.51	1.53
Weld - Slag	1.55	1.61	1.62
NT			
Weld - Machined - Sound	1.06	1.03	1.03
As Welded - Sound	1.17	1.09	1.08
Weld - Slag	1.34	1.21	1.15
Weld - Undercut	1.34	1.23	1.17

For Bending Fatigue - At Two Tensile Strength Levels

DISCONTINUITY	APPROX. % LOSS IN ENDURANCE LIMIT Q & T-122 ksi	 N & T-88 ksi
Weld - Undercut	31	22
Weld - Incomplete Penetration	22	6
Weld - Slag Inclusions	21	13
Weld - Sound - Not Machined	20	5
Weld - Sound Machined	19	2

For Torsion Fatigue - At Two Tensile Strength Levels

DISCONTINUITY	APPROX. % LOSS IN ENDURANCE LIMIT Q & T-122 ksi	 N & T-88 ksi
Weld - Undercut	36	15
Weld - Slag Inclusions, Severe	30	13
Weld - Sound - Not Machined	20	8
Weld - Lack of Penetration	15	10
Weld - Sound Machined	15	3

The loss in fatigue properties from the sound weld occurs because of the different compositions and the geometric effect of the weld reinforcement previously discussed. The sound-machined specimen loses fatigue strength only because of the difference in weld and base casting properties and composition. These specimens were heat treated after welding so any heat effects would be removed, The effect of the difference in composition is shown by the traverse of hardness readings taken across the cast steel weld deposit* area in Figures 15a and 15b (28).

Extensive experimental work has been performed on the effect that slag inclusions in welds have on weld fatigue strength (26, 29-31) Transverse butt welds were made on mild steel plate with both rutile and low hydrogen electrodes containing small discrete slag defects. The steel plate had a tensile strength of about 64.3 ksi. The decrease in fatigue strength with increasing total inclusion length, where the defect length and depth have been held constant, and the higher fatigue strength in low hydrogen welds in 1/2 inch thick plates are presented in Figures 16a and 16b (26, 29-31). These fatigue tests were conducted on two pass but welds machined flush with the surface before testing.

Additional tests were conducted with 1-1/2 inch thick butt welds. The heavier section resulted in significant residual stresses and problems with hydrogen (7, 31). The influence of a 300°F preheat and a 1200°F stress relief for 1-1/2 hours were studied. The tests were conducted on a mild steel plate with a tensile strength of 63.4 ksi. The following conclusions resulted from these tests (31).

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*Cast 8630 type steel, typically .30 C, .80 Mn, .30 Si, .45 Cr, .55 Ni, .20 Mo, SMAW, low hydrogen electrode E9018-B3 with typical .l0 C, .90 Mn, .80 Si, .15 Cr, 1.60 Ni, .35 Mo, .05 V

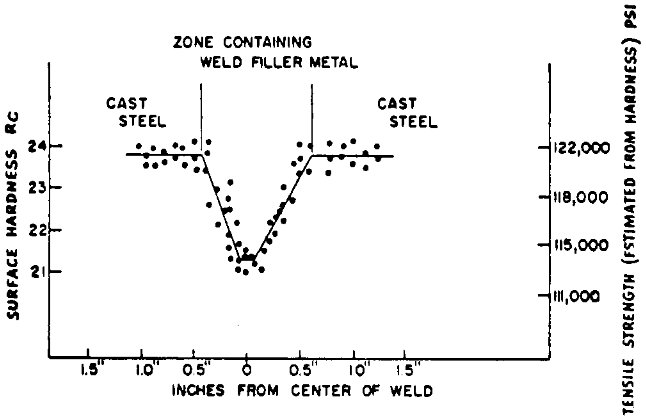

Figure 15a. Hardness profile for welded fatigue specimens after quenching and tempering (28).

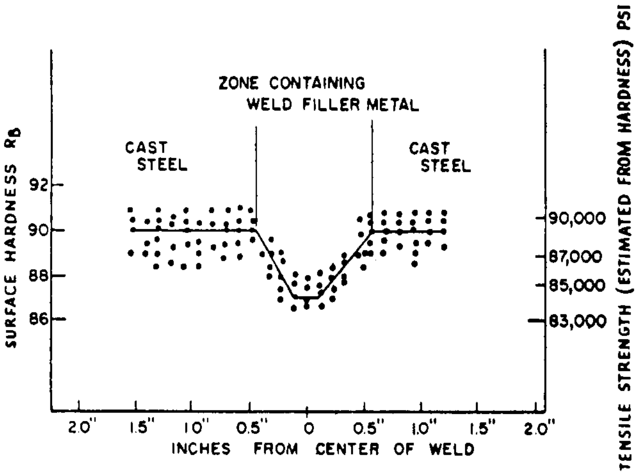

Figure 15b. Hardness profile for welded fatigue specimens after normalizing and tempering (28).

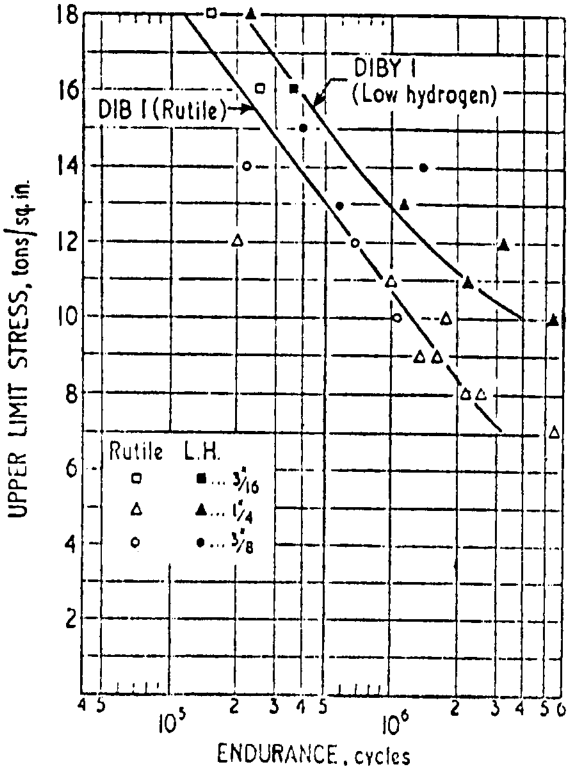

Figure 16a. Fatigue test results for rutile and low hydrogen welds (single 3/16-3/8 inch inclusions).

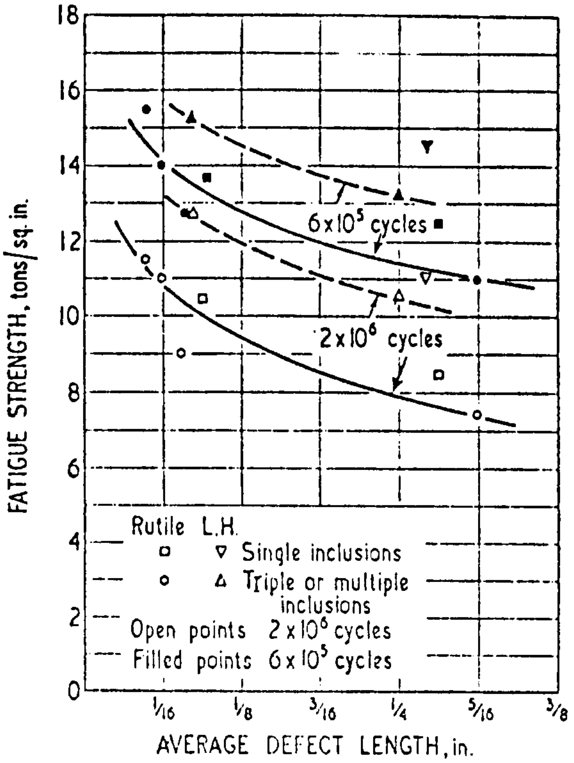

Figure 16b. Summary of fatigue tests results (25, 29-31).

(1) Discrete slag inclusions are more deleterious near the rolled plate surface because the surface is in residual tension.

(2) When the effect of discontinuities central in thickness were blanketed by compressive residual stress, large and small slag particles produced similar strengths. When these compressive stresses were relieved prior to testing, the larger inclusions resulted in a lower strength.

(3) For discontinuities central in thickness, stress relief produces an increase in strength where the slag is discrete, but a decrease with a continuous slag line. The reduction in strength, resulting from stress relief for a continuous defect central in thickness, occurs because the compressive stresses at the defect are removed by the treatment. The improvement in strength for discrete defects is that stress relief removes hydrogen.

(4) Stress relieving raises the fatigue strength of rutile weld specimens with discrete slag inclusions at their center or at their surface to a value approximately the same as that for low hydrogen weld specimens with centrally located discrete defects because of hydrogen removal. However, preheat had little effect on this low strength, mild carbon steel.

(5) The fatigue strength increased with R, the ratio of the minimum stress to the maximum stress in each stress cycle.

Lack of penetration is another weld discontinuity that reduces fatigue properties. Pulsating tension fatigue tests were performed on transverse butt welds with a lack of central penetration (32). The specimens were made from a 1/2 inch thick, 4 inch wide hot rolled and normalized mild steel plate with the analysis and mechanical properties shown below. Rutile electrodes were used and none of the specimens were pre- or post-heated. The results indicate that the fatigue strength at 2 x 10^6 cycles endurance decreases with longer lack of penetration. An approximate direct or linear relation exists between the lack of penetration (% area defective) and the reduction of fatigue strength (%). The fatigue strength at 2 x 10^6 cycles was reduced from 23 ksi with good penetration to 5,000 psi with 60% lack of penetration(17). Other work (33, 34) indicates even higher losses in fatigue strength from incomplete penetration. Reductions in endurance limit of up to 40% can occur with a 15% lack of penetration, and this can increase to over a 50% loss in fatigue strength with 30% incomplete penetration.

The large fatigue strength loss in the transverse butt welds occurs because the applied stress is transverse (normal) to the partial penetration. When the incomplete penetration lies parallel to the maximum principal stress, however, it has a negligible effect on fatigue strength. Tests on fillet welds indicate little difference in fatigue resistance between full and partial penetration longitudinal fillet welds.

Gas porosity in welds can also lower fatigue strength; the effect on low cycle fatigue strength is minor but at higher cycles, much greater reductions in fatigue strength occurs (35). The data indicate that 4-1/2% porosity in a weld can lower the endurance limit from 25 to 45%. The more severe losses occur when the porosity is present at the surface in bending and torsion. A summary of the effects of porosity on the fatigue strength at 2 x 10^6 cycles is shown from several sources (17) in Figure 17.

The marked effect of undercuts on the fatigue strength observed for welds in cast steel (28) has also been noted in other weldments(33, 36). The fatigue fracture of sound specimens involving this type of defect was always observed to originate at the root of the undercut. The fatigue strength at 2 x 10^6 cycles decreases when the depth of the undercut increases. For a steel plate with a tensile strength of 58.3 psi, a reduction of the fatigue limit was observed from 26.3 ksi to 12.8 ksi when the depth of the undercut increased from 0 to 0.35 inches, i.e., a reduction ratio of about 51% (36).

The effect of weld cracks on the fatigue strength of weldments can be expected to be marked because of the sharp stress concentration that results from this type of discontinuity. One investigation reporting on the influence of cracks was conducted on transverse butt welds containing cracks parallel to the weld direction and transverse to the applied stress. Cracks penetrating about 10% of the cross sectional area of the specimen produced considerable scatter of results; however, on the average the fatigue strength at 2 x 10^6 cycles was reduced from about 14,000 psi without cracks to only 35 to 45% of that value with cracks (17).

A broad-based investigation on the influence of a number of welding discontinuities on the fatigue behavior of welds employed by both Wohler and program tests* was conducted (37). This work illustrated the comparative effect of a number of discontinuities. These are shown below according to the relative severity with the more severe listed first: cracks, undercut, lack of fusion or penetration, slag inclusions and restart. The last discontinuity is only considered to be of minor importance.

Results from conventional fatigue tests and from program tests, which employ a statistically varying load, are displayed in tabular form as Table VII and graphical form in Figure 18 (37). This study also investigated the effect of voids and found these to be of minor significance. This is at odds with the other data on porosity and only incomplete data are presented in this latter work (37).

Welding Processes --

The effect of welding processes on the fatigue strength of welds can be influenced considerably by the shape of the weld reinforcement

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*fatigue tests involving stress amplitudes that vary in accordance with statistically determined service conditions

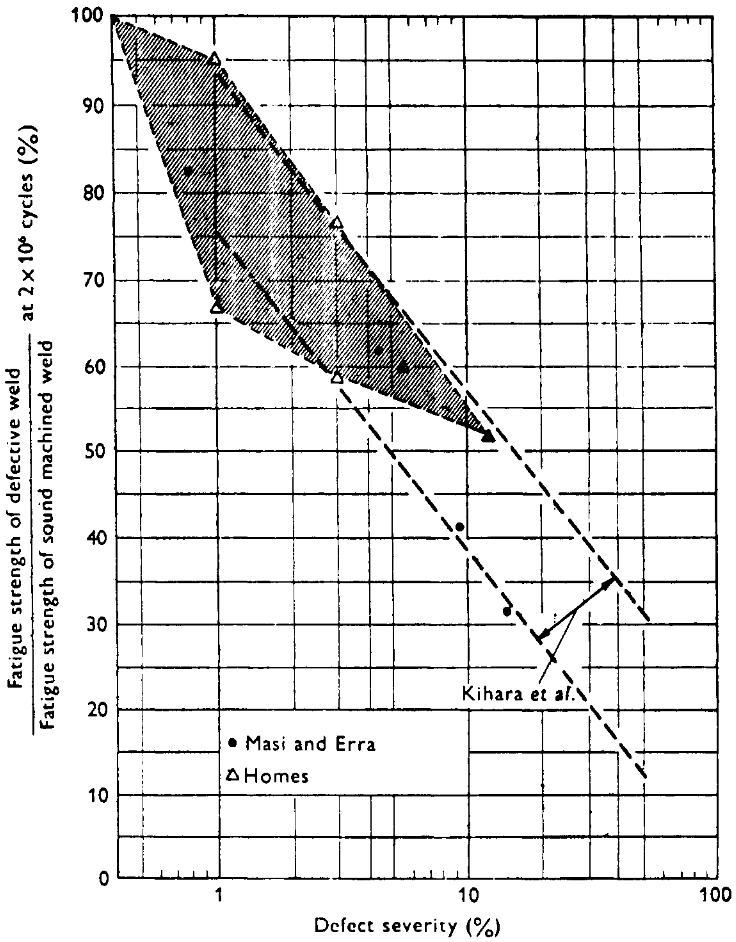


Figure 17. Plot of loss in fatigue strength at 2 x 10^6 cycles from several sources for mild steel butt welds containing different amounts of gas porosity (17).

TABLE VII: DECREASE OF FATIGUE STRENGTH UNDER CONSTANT AND PROGRAM LOAD FROM WELD DISCONTINUITIES (37).

	Without Discontinuity	Restart	Slag Inclusion	Lack of Penetration	Undercut
σ_o	29.86	-	-	-	-
σ_f	-	28.44	21.33	22.75	19.91
σ_{fr}	-	28.44	20.48	20.62	22.61
σ_f/σ_o	1.0	0.95	0.71	0.76	0.67
σ_{fr}/σ_o	1.0	0.95	0.69	0.69	0.76
$\bar{\sigma}_o$	46.93	-	-	-	-
$\bar{\sigma}_f$	-	44.08	39.11	42.66	29.86
$\bar{\sigma}_{fr}$	-	44.08	37.54	38.68	34.13
$\bar{\sigma}_f/\sigma_o$	1.0	0.94	0.83	0.81	0.64
$\bar{\sigma}_{fr}/\sigma_o$	1.0	0.94	0.80	0.82	0.73
$\bar{\sigma}_f/\sigma_o$	1.57	1.48	1.31	1.43	1.00

Stresses Shown in ksi.

All stresses are valid for 2×10^6 load cycles

Stress ratio $R = \sigma_o/\sigma_o = 0.08$

Collective corresponding normal distribution with $P = 0.5$

$\bar{\sigma}_o$ = Maximum stress for program loading

σ_o = Maximum stress for one-step test

$\sigma_f(\bar{\sigma}_f)$ = Fatigue strength with defect related to the whole specimen area

$\sigma_{fr}(\bar{\sigma}_{fr})$ = Fatigue strength reduced to the net section

σ_u = Minimum stress

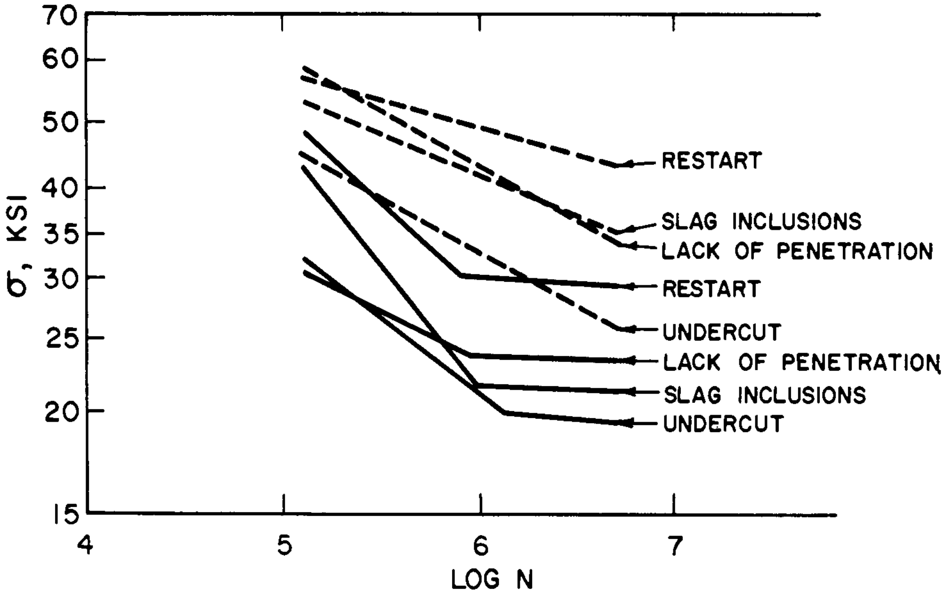


Figure 18. Results of Wohler and Program Tests with specimens containing various weld discontinuities (37).

produced by that process. One investigation of the influence of welding processes (17) conducted on transverse butt welds tested in pulsating tension concluded that automatic welds provided considerably poorer fatigue properties than those made manually. However, this difference resulted from the unfavorable reinforcement shape, as discussed under weld configuration, rather than any metallurgical difference. When the weld reinforcement was removed from automatic welds made by the submerged arc process, the fatigue strength was similar to manual welds made by the shielded metal arc method (17). These tests were stress relieved to eliminate the effect of residual stress.

Another investigation (20) concluded, based on a study of twenty-five references, that submerged arc and electroslag welding conducted automatically provided superior fatigue strength than manual welds. This improved fatigue strength was attributed to the fewer internal discontinuities and smoother weld surface obtained with the automatic processes compared to manual methods. This assumes that an unfavorable weld shape is not obtained in the automatic method. This improved fatigue strength of the automatic processes was also observed for the semi-automatic MIG process with CO_2 shielding and flux cored arc welding to a somewhat reduced extent. The electroslag process offers the potential of a considerable reduction in discontinuities with the resulting better fatigue behavior (17, 20). The differences in discontinuities can vary so widely, depending on the welding conditions, that stating specific values of fatigue strengths with the various processes is not feasible.

Residual Stress --

When a weld cools, it is restrained from contracting by the relatively cool base metal. This causes the weld to be in residual tension and the underlying casting to be in residual compression. A stress relief treatment relieves the residual tensile stresses at the toe of the weld and would be expected to increase the fatigue strength. However, it has been reported that residual stress, and hence a stress relief, has little effect on fatigue strength. In one study (38), mild steel butt welded plate was fatigue tested in pulsating tension with and without a stress relieving treatment at 1200°F. The data indicated that the stress relieving had no effect on the fatigue behavior. However, the notched fatigue strength of both a mild and medium carbon steel was reduced by the presence of a residual tensile stress when the fatigue tests were performed with a completely reversed stress cycle (39).

The relative insensitivity of weld fatigue strength to residual tension when the weld is tested in pulsating tension and the loss of fatigue strength because of residual tension when the weld is tested in a reversed stress cycle has been illustrated in other investigations(40). The influence of residual tension on weld fatigue strength is dependent on the stress ratio. The stress relief treatment improves the fatigue strength at 2×10^6 cycles as the ratio R becomes more negative or as the value of the compression stress increases. It is a general rule that no increase in fatigue strength is obtained by stress relieving structures that are subject to purely tensile loads but an improvement in fatigue strength can be obtained if the loads are alternating tension and compression.

Post Weld Treatments --

Several techniques are utilized to increase the fatigue strength of welds. These include spot heating, local compression, local heating followed by rapid quenching, prior overloading and peening (40). Grinding of the weld to reduce the stress concentration of weld configuration is also a very effective method. The applicability of most of these methods to repair or even fabricate welds in castings is limited. Spot heating and local compression are limited to welds producing localized notches such as spot welds or longitudinal weld ends (40) and are not useful for transverse welds. Local heating followed by quenching and prior overloading also are techniques limited to small weld areas and requiring special equipment; these are hardly feasible for use with castings except in unusual cases (41-43).

Peening of the weld surface with an air hammer is a technique that is adaptable to the repair and fabricated welds in castings. The improvement in the fatigue behavior obtained by peening is illustrated in Figure 19 (41). At 2 x 10^6 cycles, the fatigue strengths are 15.68 ksi and 26.88 ksi for the as-welded and peened specimens, so hammer peening increased the fatigue strength 70%. Stress relieving after peening removes most of the beneficial effect of peening. Other work shows similar improvements in fatigue strength (44).

Grinding the weld reinforcement flush to the weld surface improves the fatigue strength of welds, as discussed under weld configurations. When welds in longitudinal gussets were fully ground, the fatigue strength of mild steel fillet welds was improved from 50 to 100% (44). The effects of peening and grinding on the fatigue strength of transverse butt welds in a low carbon alloy steel with a tensile strength of 72 ksi as welded is illustrated below in Figure 20 (45). Grinding increased the endurance limit of the welded specimen by 20%, peening by 51%, and grinding plus peening by 56%. Grinding did not cause a greater increase in the endurance limit because the initial angle between the weld reinforcement and the base plate was not very sharp originally. However, the ground specimens still display a greater fatigue strength at shorter life than do the peened specimens(45).

Toughness of Welds

The toughness or impact resistance of welds in steel castings is of considerable significance for many dynamically loaded applications. The relative toughness of the cast steel or base metal is affected by many variables including: the chemical analysis, microstructural constituents, strength or hardness level and grain size. Optimum toughness is attained in cast steels (although not necessarily in welds) by a fine grained fully hardened and tempered structure with a low sulfur and phosphorus content and a carbon content below 0.36%. Considerable scatter occurs at any strength level because of the variations in these factors, but the toughness is generally reduced by higher strengths, as indicated by the Charpy V-notch values for

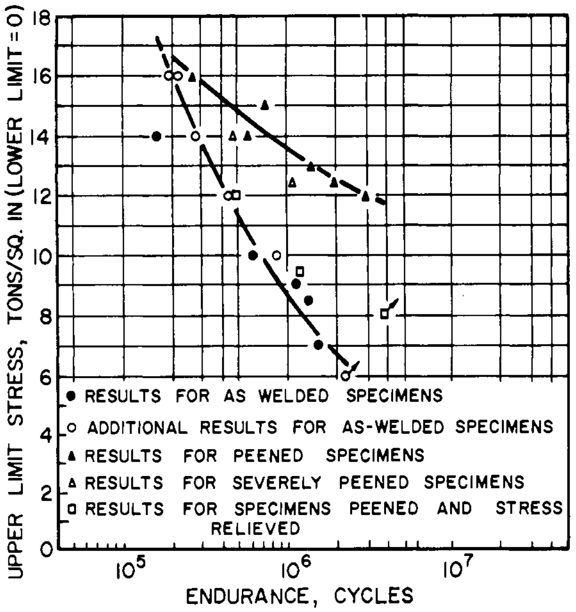


Figure 19. Fatigue behavior of as-welded, peened and peened and stress relieved transverse gusset welds tested in pulsating tension (41).

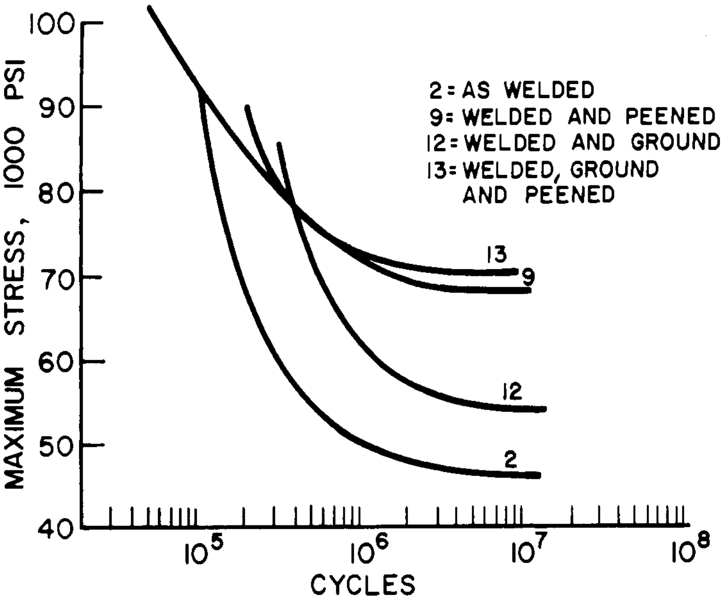


Figure 20. Effect of grinding and peening on the S-N curves of transverse butt welds (45).

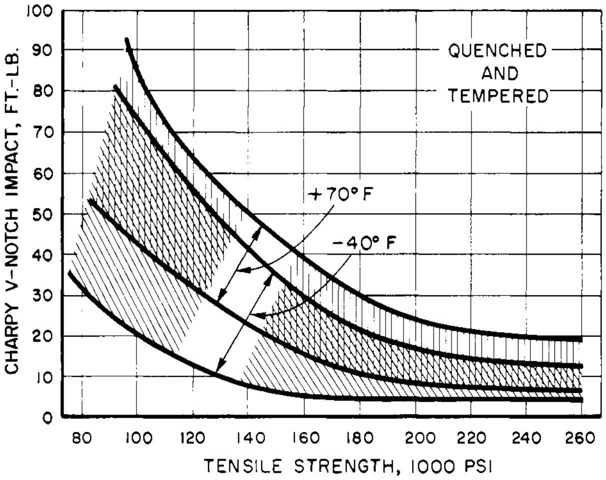


Figure 21a. Charpy V-notch impact vs. tensile strength-Cast low alloy steels in the quenched and tempered condition (tested at 70° and -40° F).

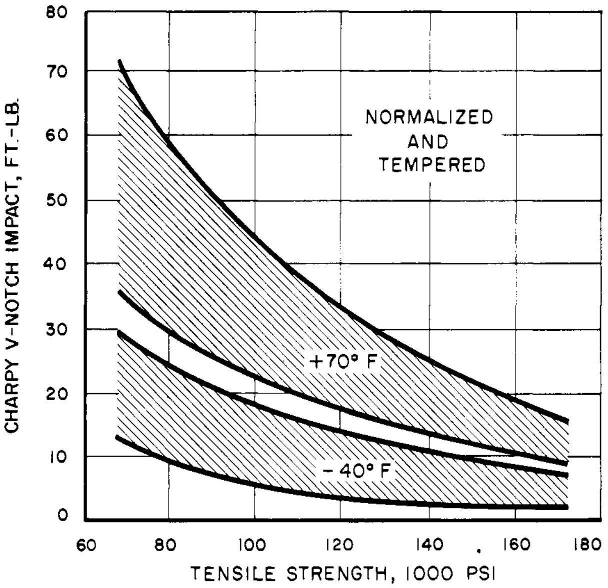


Figure 21b. Charpy V-notch impact vs. tensile strength-cast low alloy steels in the normalized and tempered condition (tested at 70° and -40° F).

tests at 70° and -40°F in Figure 21 (19). Even though these are all quenched and tempered alloy steel castings, the scatter in values at any strength level is very significant,

Welding exerts a significant influence on the toughness, particularly in cases where the castings are not completely reheat treated after welding. The variations in temperature cycles and properties demonstrated in Figure 3 produces a wide range of toughness values in the HAZ. The influence of these variations is reduced by stress relieving and further reduced by reheat treatment.

However, the following factors also affect toughness: the different welding processes such as: shielded-metal arc welding (manual-metal arc), metal-inert gas welding, flux-cored wire welding, submerged arc welding and electroslag welding; and the different features of each process including: type of electrode, heat input, welding position, heat treatment and its effect on HAZ microstructure, restraint, discontinuities and composition.

Four possible paths exist in a welded structure by which a brittle fracture could progagate. These paths shown in Figure 22 (46) include the parent plate, heat-affected zone, fusion line and the weld deposit. In a steel of high notch ductility, the plate path is eliminated as a possibility. Each region has its own characteristic fracture progagation temperature. Therefore, once fracture has been initiated, propagation occurs in the zone of lowest notch ductility. Fracture can initiate at the various types of welding discontinuities discussed in the previous section of this paper.

Types of Tests for Toughness --

Several types of tests may be employed to measure the toughness of weldments. These include: Charpy impact tests, generally V-notch; drop weight tests; dynamic tear tests; and fracture toughness tests with the associated crack opening displacement (COD) measurements. The Charpy V-notch bar has been employed for longer periods and more generally than the other types, so most of the data will refer to this. The Charpy test measurements are usually conducted over a range of temperatures to establish a brittle-ductile transition. The results are measured in terms of: fracture energy, fracture appearance (ductile or fibrous to brittle or crystalline) and the transition temperature from the brittle to ductile fracture. Many specifications require a minimum fracture energy, such as 15 ft.lbs at a given temperature.

The Charpy test does have some drawbacks. The test results are primarily qualitative and used for relative measurements of toughness rather than being directly applicable to service conditions. The test fails to differentiate between fracture initiation and propagation, although this drawback has been overcome in the instrumented Charpy test. The test is small and convenient to use but may not be applicable to larger sections.

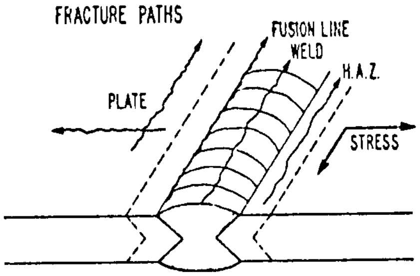


Figure 22. Possible paths of brittle fracture in weldments.

The dynamic tear test is a drop weight or pendulum test conducted on the width of a rectangular bar. This test simulates the worst condition anticipated in a structure and measures the resistance to crack propagation. It provides a much sharper indication of brittle-ductile transition when conducted over a range of temperatures than the Charpy V-notch test (47).

The drop weight test measures the ability of the steel to deform in the presence of a very sharp notch. Since the test uses an abrasive saw cut made in a brittle crack starter weld, the results are affected by the weld metal hardness and toughness of the HAZ under the starter weld. This test is also conducted over a range of temperatures to obtain the nil ductility temperature (NDT) for the steel. Both the dynamic tear and drop weight test measure the relative rather than absolute toughness of the steel.

The fracture mechanics based tests are more recent and have not been employed very widely to measure the toughness of steel welds, These tests develop a plane-strain stress intensity factor(K_{Ic}) that can be measured by calculating the stress state at the tip of a flaw in the test specimen. This property has the advantage of being a fundamental property of the material directly related to its tendency to propagate a brittle crack under stress. Accordingly, these values can be used directly in calculating material behavior under service conditions. The crack opening displacement (COD) testing extends fracture mechanics into elastic-plastic behavior. Since fracture propagation at high strain rates is much easier than fracture initiation in many of these steels under quasi-static conditions, the prevention of fracture initiation insures the avoidance of failure. The COD test permits measurements of material behavior when appreciable yielding occurs at notches or flaws before fracture. The K_{Ic} and COD measurements allow the location of the notch or crack at a specific location in the weld deposit or HAZ to measure the toughness at this location. However, the testing requirements are considerably more complex than for the Charpy, drop weight or tear tests.

Welding Processes and Electrodes --

While the level of notch toughness of the deposited weld metal will vary with the composition of the deposited metal, the cooling-conditions and the structure, a maximum attainable toughness exists for each strength level (48). This effect of tensile strength on toughness was illustrated in Figure 21 for cast steels and the same situation applies to deposited weld metal.

a. Shielded Metal Arc - Manually Applied -- Considerable information is available on the toughness of weld metal deposited from various types of standard shielded electrodes. Figure 23 (46) contains Charpy V-notch transition curves for a number of electrodes that were welded into various rolled plate steels. The class of electrode is shown for each curve with the type of steel plate in parenthesis after each one. Each curve also has a thicker black section showing the nil ductility temperature (NDT) for that steel in drop weight testing.

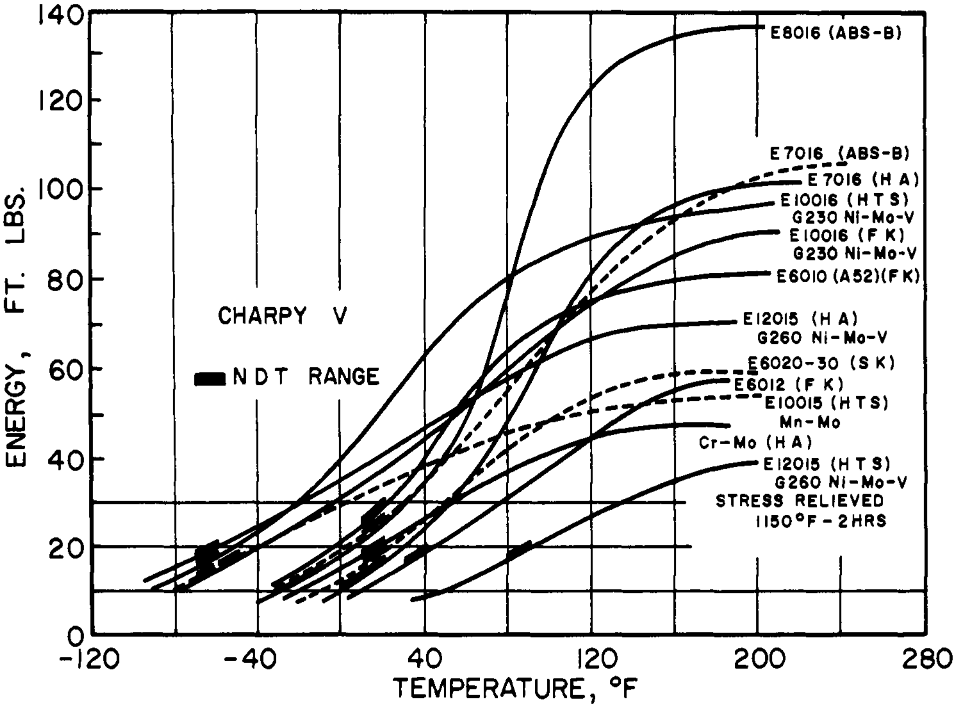


Figure 23. Charpy V-notch transition curves for various deposited weld metals. The bands indicate the nil ductility transition temperature of welds and corresponding Charpy V energies developed at these temperatures (46).

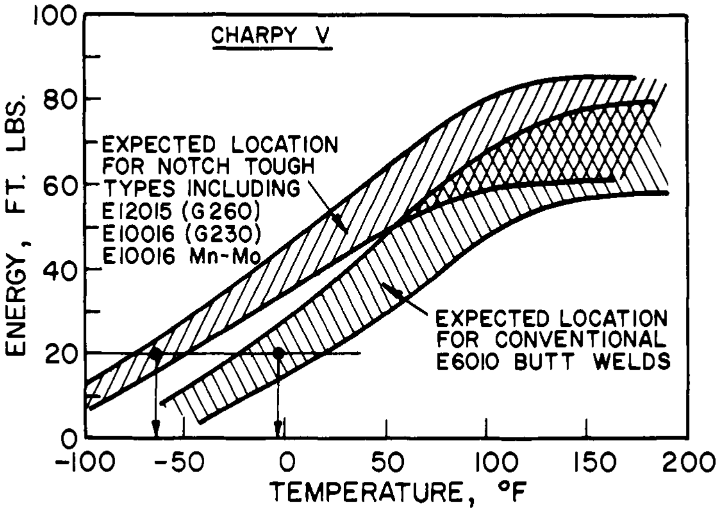


Figure 24. Typical "bands" of Charpy V-notch properties of various deposited weld steels (46).

Some general ranges or bands of Charpy V-notch transition curves for some classes of electrodes have been established as illustrated in Figure 23 (46). This plot shows the superior toughness of the EXXl5 and EXX16 low hydrogen electrodes. The low hydrogen electrodes including type EXX18 as well as EXX15 and EXX16 are strongly recommended for welds that require high toughness. It is necessary to be sure that these electrodes still have a low moisture content when employed; rebaking and using warm from the oven is recommended. While all three types of electrodes can be used, the EXX18 classes are favored because of their good welding characteristics (19). For quenched and tempered castings, E11018-G for a minimum yield strength of 100 ksi and E11018-G or E12018-GF for a minimum yield strength of 110 ksi are preferred. The typical Charpy-V-notch transition curves for E11016 and E11018 weld deposits shown in Figure 25 (19) illustrate this toughness behavior. The data below this curve shows the composition and tensile properties of the deposited weld metal.

The reasons that low hydrogen electrodes are about the only type used for welding high-strength notch-tough steels are: less difficulty with underbead cracking that could occur on the alloy grades of steel used for toughness; better impact properties, as shown in Figures 24 and 25; and the basic mineral coating makes it feasible to add carbon and other alloying elements to produce weld metals of varying compositions and strengths (48). The superior impact properties of weld metal deposited from low hydrogen electrodes have also been shown in other work (49). The toughness of low hydrogen shielded metal arc deposits with a Ni-Cr-Mo-V composition and tensile properties of about 140 ksi yield strength, 149 ksi tensile strength, 16% elongation and 60% reduction in area are shown for both Charpy V-notch and dynamic tear tests in Figure 26 (50).

A comparison of the Charpy V-notch toughness of EXX18 manual-metal arc weld deposits made with high-purity core wire to other weld processes is shown in Figure 27 (51). A vacuum-melted core wire used in the manual-metal arc process produced welds tougher than submerged arc or manual-metal arc (low purity core wire) weld, but less tough than gas-tungsten arc or gas-metal arc welds.

The effect of the flux composition in manual electrodes on the toughness has also been studied (49, 52). This work indicated that low hydrogen basic coated (lime) electrodes provided better toughness, as measured by both Charpy and COD tests, than cellulosic or rutile (titania) coatings. The basic coatings resulted in a lower non-metallic inclusion distribution because of their fluxing action.

b. Flux Cored Electrodes -- The use of flux cored tubular rods with CO_2 shielding is increasing because of the high deposition rates and lower cost (19). No industry-wide standards are available for welding wire in this process. The weld metal has good toughness at low temperatures along with a minimum yield strength of 100 ksi as indicated by the data in Table VIII (19) from three producers.

Some investigations have shown low toughness in flux cored wire

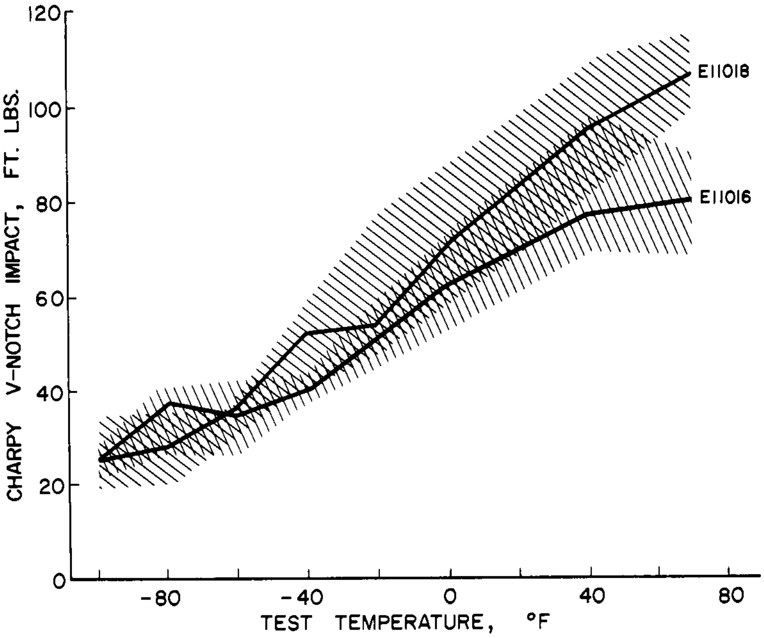


Figure 25. Charpy V-notch transition curves, compositions and tensile properties of weld metal deposited from E11016 and E11018 electrodes (19).

	E11018	E11016		E11018	E11016
%C	0.079	0.046	U.T.S., ksi	111.5	114.5
%Mn	1.36	1.18	Y.S., ksi	99.5	104.0
%P	0.017	0.013	El. % in 2"	25.0	20.5
%S	0.017	0.018	R.A. %	69.0	61.1
%Si	0.36	0.38			
%Cr	0.35	0.69			
%Mo	0.47	0.37			
%Ni	1.55	3.14			

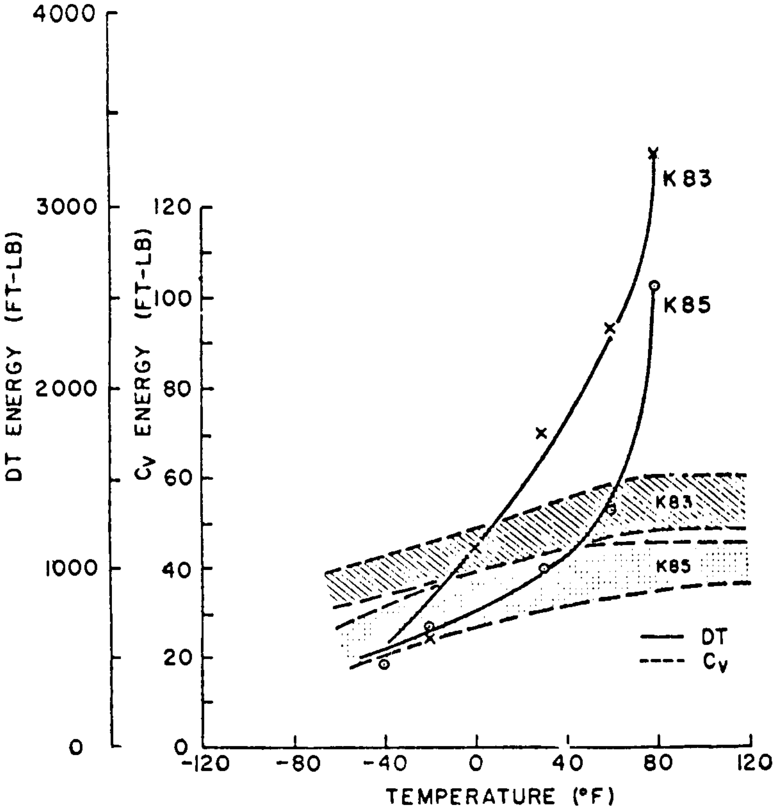


Figure 26. Dynamic tear and Charpy V-notch test temperature transitions for shielded metal-arc alloy weld metals (50).

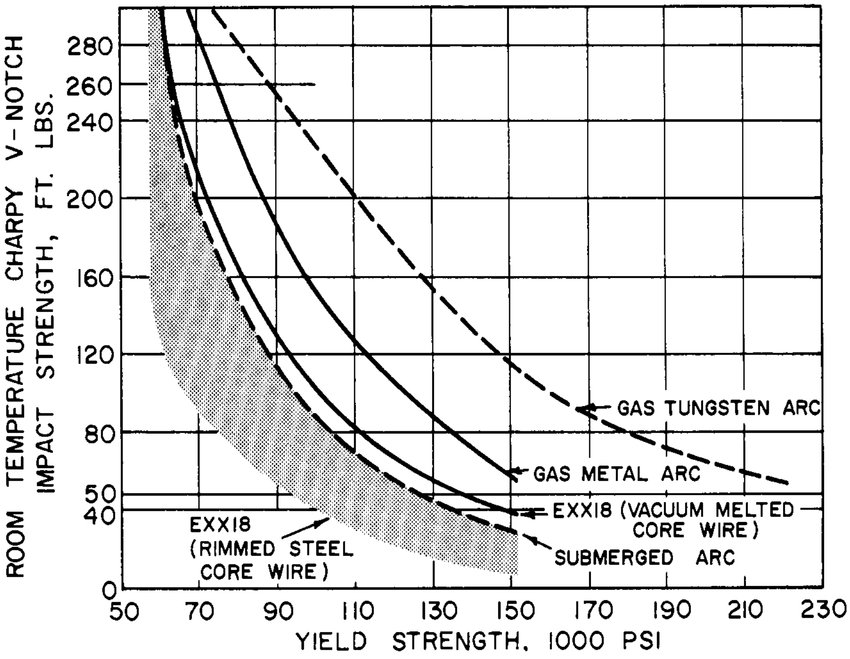


Figure 27. Comparative room temperature Charpy V-notch toughness of different electrodes in SMAW with GMAW (MIG), GTAW (TIG) and submerged arc welding processes (SAW) (51).

TABLE VIII: CHARPY V-NOTCH AND TENSILE PROPERTIES OF FLUX CORED TUBULAR RODS USED WITH CO_2 SHIELDING(19).

Manufacturer	Airco DCS	Airco DCS	McKay Speed Alloy 110		National Cylinder Gas Dual Shield T-15	
Condition of Weld	As Welded	As Welded	As Welded	Stress Relieved at 1150°F	As Welded	Stress Relieved at 1050°F
Tensile Strength, ksi	115.9	108.7	110	113.3	119	121
Yield Strength, ksi	101.8	100.5	97.8	102.7	107.5	109
Elongation, % in 2 in.	22	22.5	20	22.5	20	20
Reduction of Area, %	60.1	61.1	41.9	55.8	45	50
Charpy V-Notch Impact, ft.lb.						
72/75°F			56	54	47	40
+10°F	50(48)*	47(44)				
-40°F			42	38		
-50°F	38(28)	37(28)				
-60°F					27	22
-75°F					24	19
-95°F	27(17)	28(18)				
-100°F			32	25		

*Values in Parenthesis Show Charpy V-Notch Impact After Stress Relieving.

to be associated with unconsumed deoxidizing agents in the weld deposit (53, 54). The presence of silicon, aluminum and titanium increased the transition temperature for the Charpy test. A regression analysis equation was developed that shows the relative effect of each of these elements on this transition temperature (53, 54). Titanium reduced the toughness the most, followed by silicon and then aluminum for each percent of the element present.

c. Submerged Arc -- The submerged arc welding process is characterized by larger size weld deposits and higher heat inputs than the manual shielded arc or flux cored wire. The notch toughness is lower for the larger, fewer passes than obtained with multilayer procedures. A comparison of the Charpy V-notch toughness obtained by single and multipass deposits in carbon steel plates is indicated in Figure 28 (46). Other investigators have shown that as the number of passes in submerged arc welds increased, the toughness also improved(55, 56). The heat input decreased with an increasing number of passes because of decreased amperage and faster travel speed. The yield and tensile strength increased from about 63 to 90 ksi and 80 ksi to 100 ksi, respectively, with a decrease in elongation from 20 to 28% as the passes increased from 12 to 110. Table IX (55) indicates the change in the toughness as determined by the Charpy V-notch specimen over this number of passes. Charpy transition temperatures were lowered as the number of passes increased up to 66-69; little effect occurred with additional passes. This effect is attributed to more uniformity and weld refinement with increasing numbers of passes. Other work obtained mixed results on the influence of the number of passes on the properties (48).

Research has been conducted on both the filler rod composition and flux used for HY-80 steel welds made with the submerged arc process (48). By improving the cleanliness and reducing the oxygen content of the filler rod and weld deposit, the toughness as measured by the Charpy test was improved significantly. The fracture energies have increased from 28 to 62 ft.lbs at room temperature and 21 to 56 ft.lbs at -40°F with the better weld metal and flux.

The flux composition also exerts a marked influence on the toughness of submerged arc welds (48, 57, 58). The effect of a variation in flux composition from acid to fully basic on the analysis, tensile properties and Charpy V-notch transition curve is shown in Table X and Figure 29 (57). The influence of slag basicity on the transition curve obtained with a Charpy V-notch impact test is also shown in Figure 30 (58) from another investigation. It is evident from this study that higher flux basicities produce improved toughness and less oxygen in the deposited weld metal. Similar behavior has been observed by other investigators (48).

Acid fluxes produced a viscous slag which was easily entrapped during solidification, forming large globular inclusions and brittle grain boundary films. Basic fluxes produced a very fluid slag which was not entrapped during solidification, and lowered the sulfur and oxygen contents in the weld. Therefore, the basic flux increased toughness by improving steel cleanliness through decreased size and amount of inclusions.

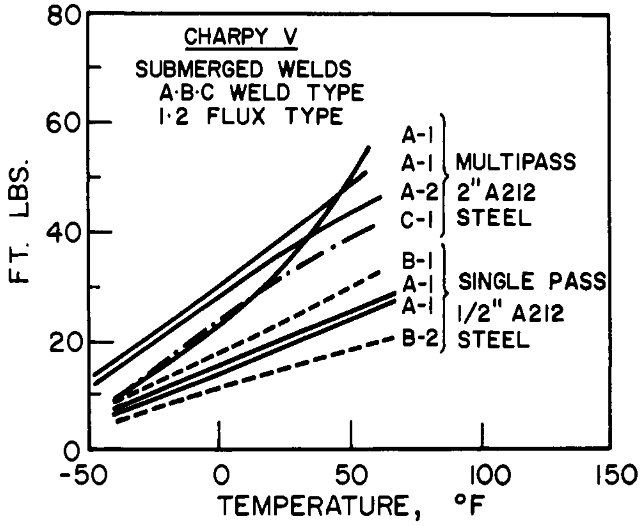


Figure 28. Charpy V-notch curves for single and multipass submerged arc welds.

TABLE IX: INFLUENCE OF NUMBER OF PASSES ON THE TOUGHNESS OF SUBMERGED ARC WELDING (55).

Transition Criteria	Energy Absorption			Fracture Appearance	
Number of Passes	15 ft/lb	20 ft/lb	35 ft/lb	20% Fiber	50% Fiber
12-16	-41°C	-33°C	-17°C	- 8°C	+25°C
35-37	-37°C	-29°C	-15°C	-14°C	+13°C
66-69	Below -60°C	-59°C	-29°C	- 7°C	+14°C
108-110	Below -60°C	-60°C	-11°C	-21°C	+12°C

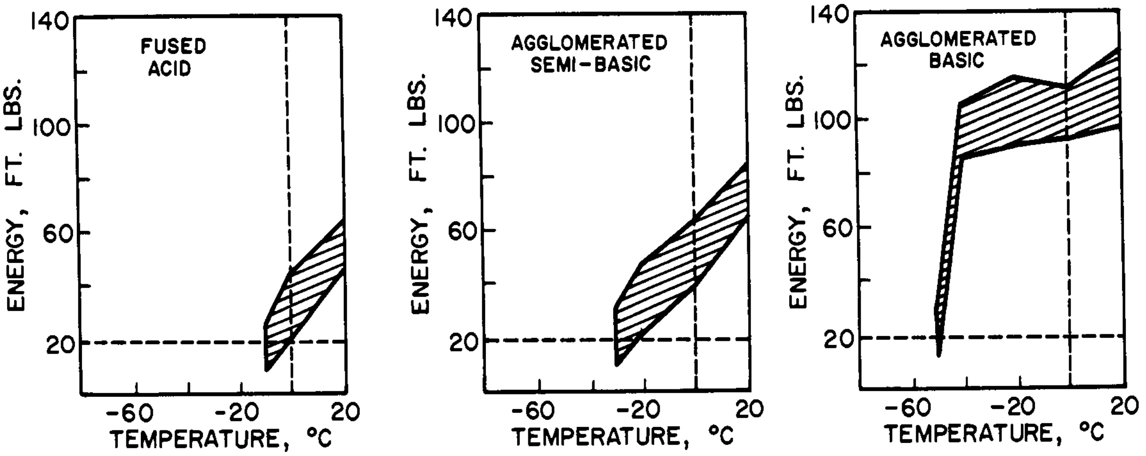


Figure 29. Charpy V-notch transition behavior of submerged arc welds made with different fluxes (57).

Conversion:	°C	-60	-40	-20	0	20
	°F	-76	-40	- 4	32	68

TABLE X: CHEMICAL ANALYSIS AND MECHANICAL PROPERTIES OBTAINED WITH DIFFERENT FLUXES FOR SUBMERGED ARC WELDS (57).

Flux Type	Analysis Obtained on Weld Deposits				
	Carbon,%	Manganese,%	Silicon,%	Sulfur,%	Phosphorus,%
Fused Acid	0.03	1.32	0.48	0.021	0.024
Agglomerated Semi-Basic	0.05	1.09	0.42	0.19	0.025
Agglomerated Basic	0.075	1.13	0.29	0.11	0.021

TABLE X (Continued):

Flux Type	Mechanical Properties		
	Yield Strength (ksi)	UTS (ksi)	Elongation (%)
Fused Acid	49.6	62.8	28
Agglomerated Semi-Basic	52.7	67.6	31
Agglomerated Basic	58.9	76.4	34

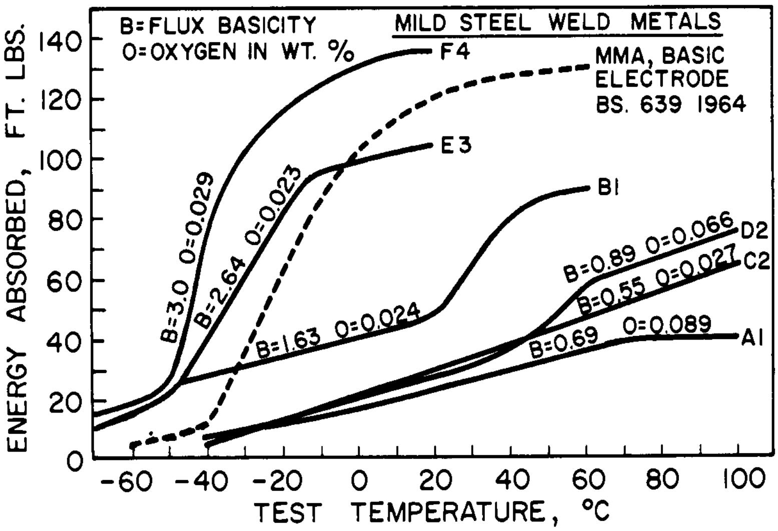


Figure 30. Charpy V-notch impact transition curves of mild steel weld metals deposited by submerged arc process showing the effect of flux basicity.

Conversion:	°C	-60	-40	-20	0	20	40	60	80
	°F	-76	-40	- 4	32	68	104	140	176

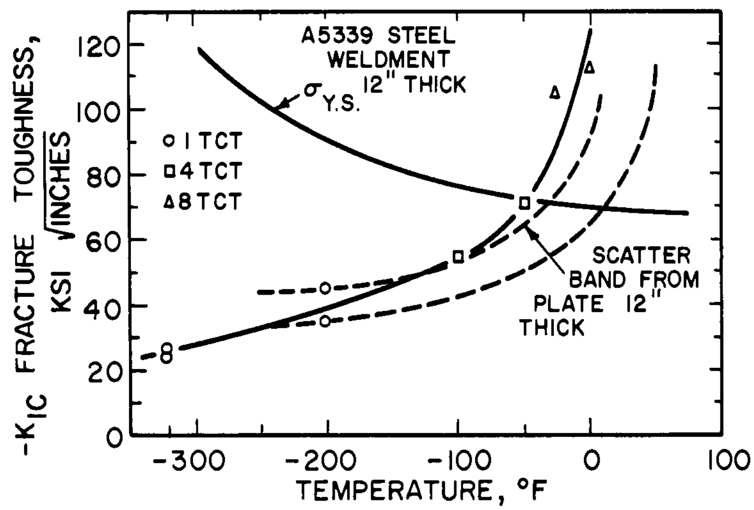


Figure 31. Temperature dependence of the K_{Ic} fracture toughness of 12 inch thick weld deposit in A533 Grade B, Class 1 steel. (weld deposits by submerged arc process) (59).

Submerged arc welding is capable of providing good toughness in heavy plate sections with the appropriate technique. Welds were made in a 12 inch thick plate of a composition approximately 0.20% C, 1.4% Mn, 0.25% Si, 0.6% Ni and 0.5% Mo with room temperature tensile properties of 85 ksi tensile strength, 70 ksi yield strength, 25% elongation and 63% reduction of area (59). The multiple pass welds were deposited by the submerged arc process using 1/8 inch electrodes, 600 amps, 32 volts at a travel speed of 20 inches per minute. A preheat of 300°F, interpass temperature of 500°F and post weld treatment of 1100-1150°F for 12 hours was employed. The resulting toughness is shown by the K_{Ic} values at various temperatures in Figure 31 (59).

d. Gas Shielded Arc -- The gas shielded metal arc welding process utilizes CO_2, argon and other gas combinations for shielding atmospheres. This process has the capability of providing good weld toughness as shown in Figure 27. The CO_2 shield, because of its lower cost, is more widely employed, although combinations of CO_2-O_2-argon, argon-O_2 and argon-CO_2-O_2-N_2 have been used (48). Figure 32 (60) illustrates the Charpy V-notch transition curves obtained in the as-welded and stress relieved condition on welds made in a low carbon steel plate with a CO_2 atmosphere shield. This figure also indicates the toughness obtained with different methods of arc transfer. Welds made by short-circuiting arc transfer had better toughness than those produced with a spray type arc (52, 61), particularly at low temperatures. However, other data (48) shows somewhat better toughness for the spray arc compared to the short-circuiting arc. Response to stress relief was most marked in welds made with silicon-manganese deoxidized wires; the Charpy 15 ft.lb transition temperature was depressed to below -58°F (wires C, D, E and F).

It has also been reported (61) that higher voltages and currents decreased notch toughness in both single and multipass welds. Multipass welds have the advantages of smaller weld pools with faster solidification and providing heat treatment for previous deposited beads. Both of these factors lead to refined grain size in the weld deposit. Welds in high toughness steel (HY 130) are reported (50) to provide Charpy V-notch fracture transition temperatures in the range of 0 to -50°F when produced with CO_2 gas shielding, The toughness attained increases with higher amounts of argon in the shielding gas. Changing the shielding gas from 100% CO_2 to 80% argon, 20% to 100% argon in welding a low carbon alloy steel plate increased the Charpy V-notch fracture energy at -40°F from 12 to 23 to 32 ft.lbs, respectively (48).

e. Electroslag -- The electroslag process is most suitable to heavier sections, at least over a 2 inch thickness. The toughness in both the deposited weld metal and in the HAZ is very low (48, 62, 63). This effect is attributed to the large heat input in the process with the consequential large grains in both the deposited weld metal and HAZ. Efforts have been made to improve the toughness of the deposited weld metal by alloying and vibration during welding (48), but this does not solve the problem of the coarse grains in the HAZ. The data plotted in Figure 33 (48) show the Charpy V-notch transition curves plus selected temperature tests for weld metal as deposited by the electroslag process.

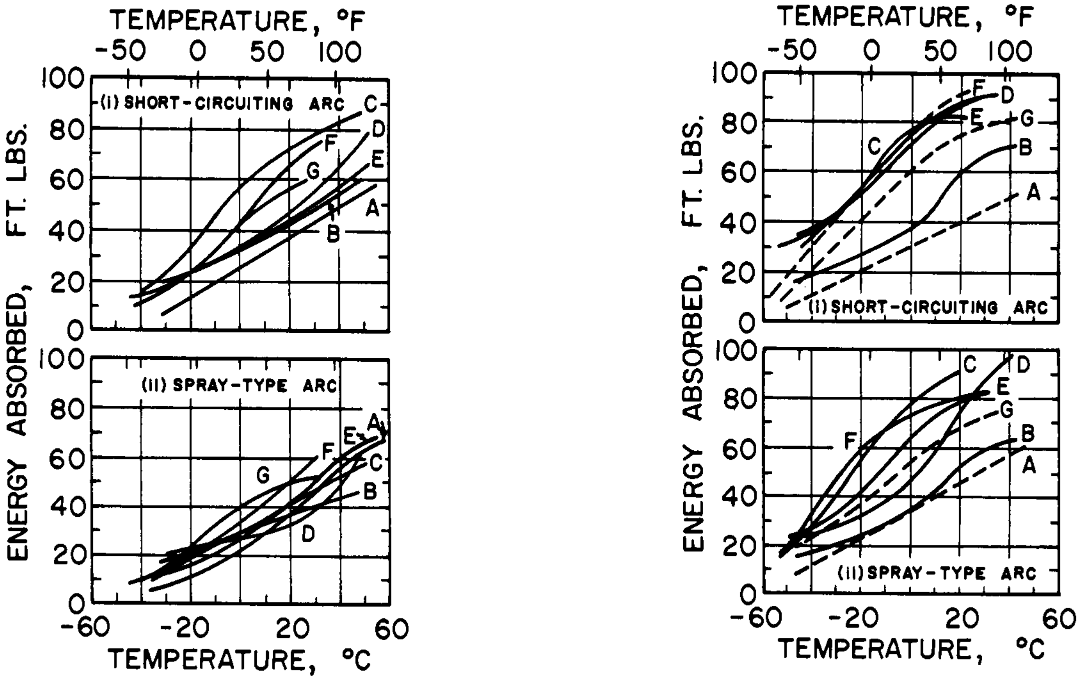


Figure 32. Charpy V-notch transition curves for welds produced by short-circuiting and spray type of arc using CO_2 gas shielded arc welding process (61). As-welded (left) and stress relieved (right).

Conversion:							
°C	-60	-40	-20	0	20	40	60
°F	-76	-40	- 4	32	68	104	140

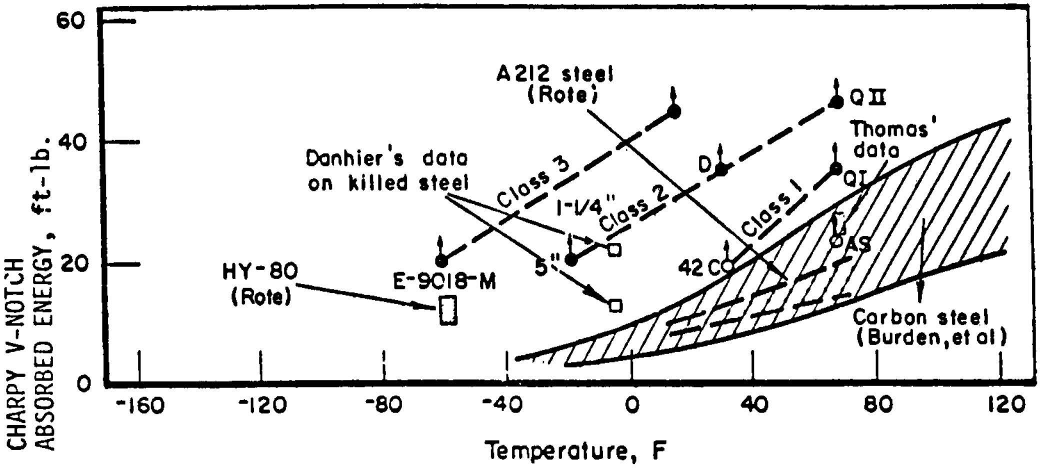


Figure 33. Charpy V-notch toughness of as-welded steels deposited by the electroslag process (48).

Figure 34 (48) illustrates the improvements that can be obtained in the Charpy V-notch toughness by stress relieving, normalizing and quenching and tempering. The heat treating processes, particularly normalizing and stress relieving (normalizing and tempering), improves the as-welded toughness considerably, but it still is significantly lower than obtained by the other arc welding methods.

Toughness in HAZ --

The HAZ, as discussed in the section on weldability, contains a number of regions that have been heated to various temperatures above and below the critical temperature ranges of the steel during fusion welding. The grain coarsened region in this HAZ (location 1 in Figure 3) is usually considered to be the area of maximum embrittlement. The loss of toughness of the HAZ compared to the weld metal and base plate has been demonstrated by comparing the Charpy V-notch impact test results of these with the test results on a specimen that consists of a combination of HAZ and weld metal, as illustrated in Figure 35 (64) for a low alloy steel. It is apparent that the combined HAZ and weld metal has a lower toughness than either the weld or base metal. The coarse grain size of the portion of the HAZ near the fusion zone has the lowest toughness as indicated by the COD measurements on grain coarsened and grain refined HAZ steel compared to the base plate for a low alloy steel in Figure 36 (68).

The toughness of the HAZ has also been studied by investigators using specimens that had been heat treated through cycles simulating the time-temperature history of the HAZ. This method allows the testing of larger specimens because it avoids the rapidly changing structures that occur over small distances in actual weld HAZ. These test results (65, 66, 69, 70) also demonstrated, using COD tests, that the grain coarsened region of the HAZ had the lowest toughness.

The notch toughness of the HAZ in carbon steels has been studied (65, 67) with about the same results as those obtained for low alloy steels. Maximum hardness and the highest Charpy V-notch transition temperatures measured by both fracture appearance and energy transition occurred in the portion of the HAZ heated to the highest temperature. One investigation (67) also showed a region of embrittlement in carbon steels located in the thermally strained zone just outside of the HAZ as a result of strain aging. This embrittlement would not occur in an aluminum deoxidized steel.

The effect of various thermal cycles on the toughness of the HAZ, as well as the effect of post-weld heat treatments, has been studied extensively on specimens of HY 80 subjected to simulated thermal cycles (66). The thermal cycles were selected to duplicate the following conditions in the HAZ for single pass and multiple pass welds: grain coarsening, grain refining and the intercritical temperature range between the A_1 and A_3 temperatures. These thermal cycles involved heating the grain coarsened specimens to 2327°F, the grain refined specimens to l705°F and the intercritical specimens to 1410°F. The Charpy V-notch transition curve indicated that in the as-welded condition, the grain coarsened area had the lowest toughness followed by

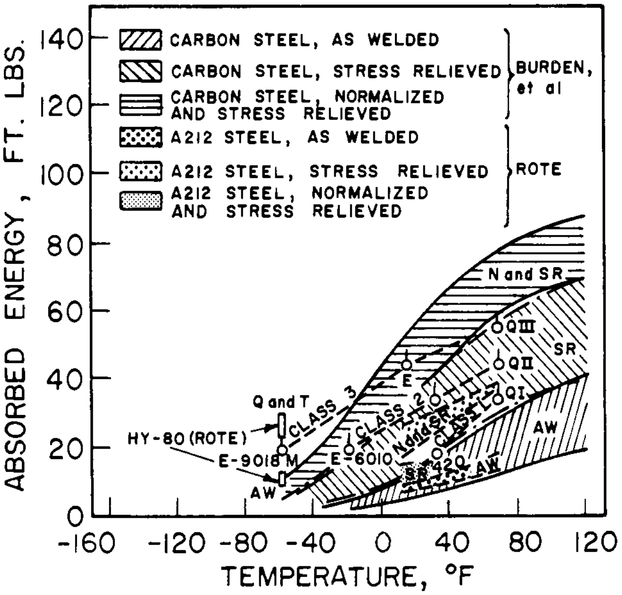


Figure 34. Improvement in Charpy V-notch toughness of electroslag-deposited steels obtained by heat treatment (48).

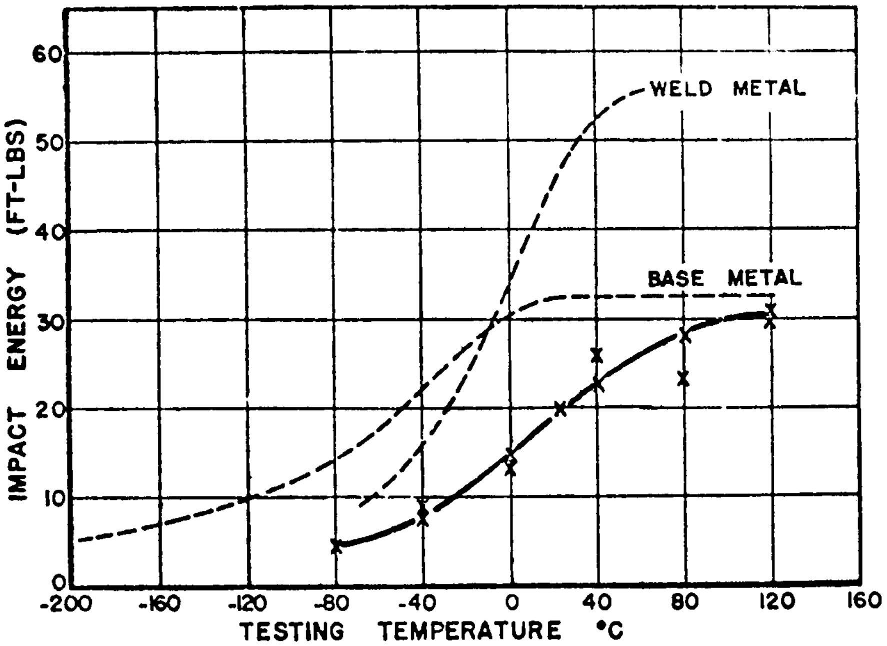


Figure 35. Transition curves for base metal E12015 weld metal and the composite weld joint (64).

Conversion:	°C	-200	-160	-120	-80	0	80	160
	°F	-328	-256	-184	-112	32	176	320

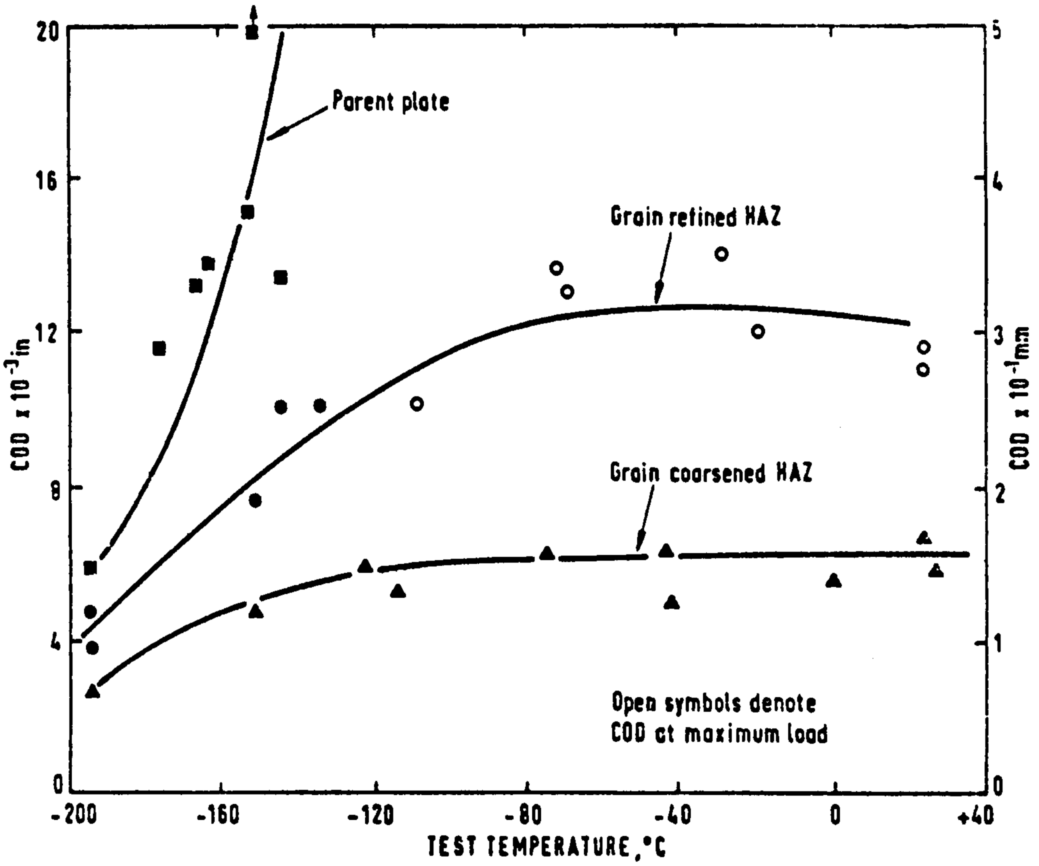


Figure 36. Fracture toughness tests on HAZ regions and parent plate of a Ni-Cr-Mo quenched and tempered steel. Specimen size 10 mm square x 55 mm. (68).
Conversion: °C -200 -160 -120 -80 0 40
°F -328 -256 -184 -112 32 104

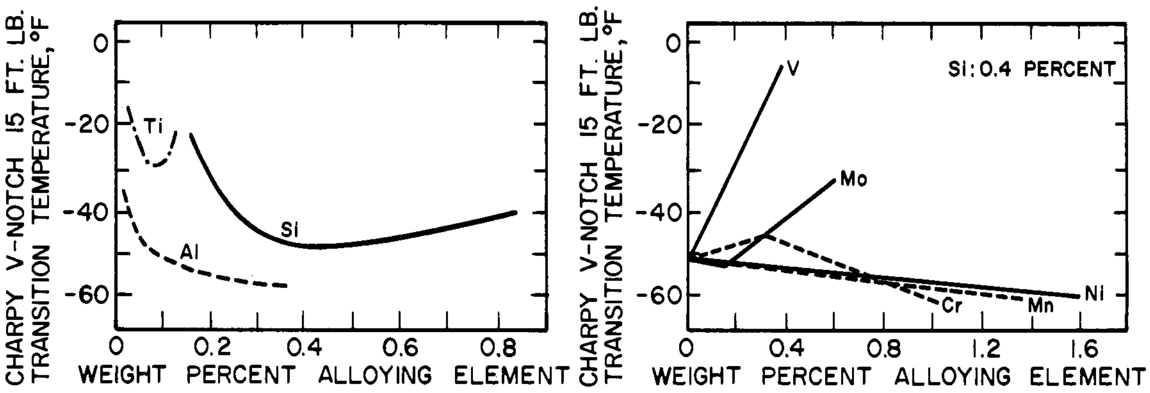


Figure 37. Effects of alloying elements on the notch toughness of weld metals deposited from covered electrodes (71).

the grain refined steel; the intercritical specimens had the highest toughness, and it was equivalent to the base metal as welded. Postweld heat treatments improved the toughness of the grain coarsened steel significantly, but it did not reach the level of toughness of the HY 80 plate. The postweld treatment was successful in returning the grain refined portion of the HAZ to the tougheess of the original plate. These studies also indicated (67) that the embrittlement of the HAZ was greater in quenched and tempered than in normalized and tempered steels.

Effects of Various Factors on Toughness --

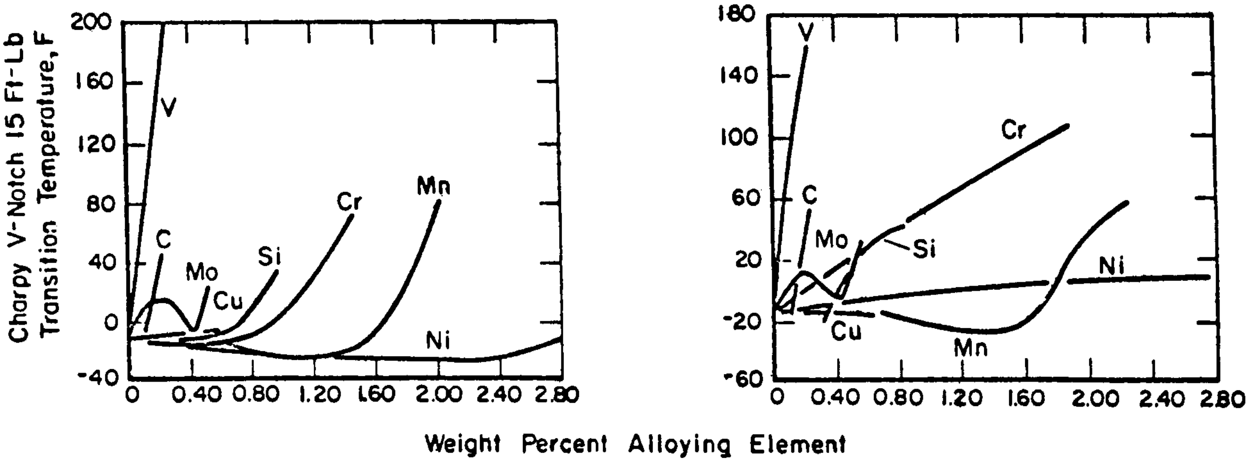
a. Chemical Composition -- The composition of most of the weld deposits used for joining carbon, low and medium alloy steels are low carbon, low alloy ferritic steels. They contain less than 0.2% carbon and combinations of manganese, nickel, chromium, vanadium and molybdenum in mounts less than 5%. Other significant elements present, although not intentionally added, are sulfur, phosphorus, oxygen and nitrogen. The effects of alloying elements on the notch toughness of covered electrodes with a lime-fluoride type coating was determined by adding various alloying elements to the coating on a low carbon, low manganese core wire (71). The multilayer welds were deposited by the downhand method. The results of this work are summarized in Figure 37 (71). This figure demonstrates the Charpy V-notch 15 ft.lb transition temperature of the weld metal with different amounts of the alloys present.

A statistical study was conducted of the notch toughness of multilayer weld metals deposited with commercial and experimental basic-type electrodes (72). The following equation was formulated to predict the influence of alloying elements (in percent) and grain diameter on the 15 ft.lb Charpy V-notch transition temperature (Tr_{15}). This formula is capable of predicting the experimentally determined 15°F transition temperature fairly closely (72).

$$
\begin{aligned}
Tr_{15}(°F) = {} & 436\,C - 54\,Mn + 14\,Si + 286\,P + 819\,S - 61\,Cu \\
& - 29\,Ni + 13\,Cr + 23\,Mo + 355\,V - 112\,Al \\
& + 1138\,N + 380\,O + 1.08\,(d \times 10^{4}) - 203 \pm 22
\end{aligned}
$$

where d is grain diameter, inch (ranged 3.1 to 8.7 x 10^{-4} inch), and the elements were: C - 0.03-0.11; Mn - 0.2-0.16; Si - 0.5-1.2; P - 0.004-0.17; S - 0.006-0.11; Cu - 0.05-0.3; Ni - 0.05-1.4; Cr - 0.05-2.6; Mo $\leq$1.2; V $\leq$0.31; Al $\leq$0.36; N - 0.004-0.02; O - 0.007-0.19.

The effect of alloying elements on the notch toughness of submerged arc and inert-gas deposited metal has also been measured by adding eight common alloying elements singly to the weld deposit (73). The effect of these elements on the Charpy V-notch 15 ft.lb transition temperature is indicated in Figure 38 (73). The chemical compositions of the base metal, welding wire and base weld metal are listed below the figure.



a. Submerged-Arc Deposited Metals

b. Inert-Gas Metal-Arc Deposited Metals

Figure 38. Effects of alloy additions on notch toughness of weld metals made with submerged arc welding and inert gas metal arc welding (73).

	Submerged-Arc			Inert-Gas Metal-Arc		
	Base Metal	Welding Wire(a)	Base Weld Metal	Base Metal	Welding Wire(b)	Base Weld Metal
Carbon	0.16	0.10	0.12	0.16	0.09	0.12
Manganese	0.47	0.41	0.60	0.47	1.27	0.70
Silicon	0.10	0.01	0.30	0.10	0.54	0.35
Phosphorus	0.014	0.011	0.013	0.014	0.015	0.013
Sulfur	0.035	0.029	0.033	0.035	0.022	0.028
Copper	...	...	0.04	...	...	0.20

(a) Bare wire.
(b) Copper-coated wire.

The effect of alloying elements on the base steel, except for impurity elements, have their primary significance in how they influence the hardness developed in the HAZ by the welding process. The data in Figure 39 (74) indicates that the Charpy V-notch 40 ft.lb transition temperature in the stress relieved HAZ of a nickel-chromium-molybdenum-vanadium steel was raised by all elements except nickel.

b. Individual Elements -- Carbon is kept low in weld deposits to maintain higher toughness, as shown in Figure 38, although this element is necessary for strength and should not be reduced to the lowest possible percentage. Carbon up to 0.20% increases the hardness and strength of the deposit but lowers the toughness (73). Carbon contents from 0.05 to 0.10% are suggested to provide higher strength without cracking problems (49, 75).

According to the data in Figure 37, manganese provides strength with no loss in toughness up to about 1.4% for covered electrodes. Manganese contents up to l.5% also improve the toughness slightly in both submerged arc and gas-metal arc deposited weld metals, as demonstrated by the data in Figure 38. Above 1.5% Mn, however, the transition temperature increases significantly. Other work (49) indicates that manganese up to 2% improves toughness; higher than 2% Mn has been reported elsewhere to lower toughness (57). Manganese has been attributed by some investigators (76) to contribute to a loss of toughness by increasing temper embrittlement in a 2-1/2 Cr- 1 Mo shielded arc deposit, and it was recommended in this work (76) that this element be maintained at low levels. Manganese in the base steel HAZ reduces toughness by its hardening effect, as shown in Figure 39.

Silicon is added to weld metal as a deoxidizing rather than alloying element, and a range of 0.25 to 0.40% Si has been recommended to avoid porosity in a 140 ksi yield strength weld deposit (75). An improvement in toughness was observed in shielded metal arc deposits with a silicon increase from 0.2 to 0.4% as shown in Figure 39. An increase in the transition temperature of inert gas shielded weld deposits was noted for increasing silicon from 0.35 to 0.8%. The toughness of submerged arc weld deposits increased with silicon contents of over 0.5% (see Figure 38). Silicon has also been considered to be a major contributor to temper brittleness in 2-1/4 Cr - 1 Mo welds (76).

A relatively high ratio of Mn:Si has been proposed in welding wire so that the FeO-MnO-SiO_2 liquid solution formed by reactions with oxygen will be unsaturated with SiO_2. This composition is reported to allow the deoxidation products to float to the surface, leaving clean weld metal (48).

Chromium is not as beneficial as manganese, molybdenum or nickel in improving the toughness of weld deposits (49). The data for shielded metal arc welds shown in Figure 37 indicates that the Charpy V-notch 15 ft.lb transition temperature of covered electrode deposited weld metal increased slightly with the addition of chromium up to 0.4% and then decreased as the chromium content increased to about 1%. On the other hand, the data in Figure 38 shows that increasing chromium content up to 1.6% caused a progressive increase in the transition temperature of submerged arc and inert gas metal arc deposited weld metal. The thermal cycles of the welding process affects the response of chromium considerably. When 2 1/2% chromium is present in weld metal, a post heating temperature of 1300°F considerably improves the toughness (48). Chromium in the base metal can reduce toughness as illustrated in Figure 39.

Nickel produces a moderate strengthening of weld deposits with a slight improvement or at least no significant decrease in toughness. The data of Figure 37 show a slight decrease in the 15 ft.lb Charpy V-notch transition temperature with increasing nickel up to 1.6% in shielded metal. arc welds. Figure 38 indicates a small decrease in this transition temperature up to 2.5% nickel in submerged arc welds but a slight increase for all nickel contents up to 2.7% in inert gas welds. Nickel does improve the toughness of the HAZ when present in the base plate, as indicated in Figure 39.

Molybdenum raises the strength but also decreases the toughness of weld deposits, although its effect varies with the welding process. Additions of over 0.2% increase the 15 ft .lb transition temperature of shielded arc welds (see Figure 37). In submerged arc and inert gas metal welds, molybdenum up to 0.20% increases this transition temperature. The transition temperature decreased slightly from 0.20 to 0.45% Mo and then increased again, as indicated in Figure 38. Molybdenum also decreased the toughness of the HAZ, as demonstrated in Figure 39.

Vanadium markedly lowers the toughness of all types of welds and the HAZ, as demonstrated in Figures 37-39. This element sharply increases the susceptibility to stress relief embrittlement (48,49,75).

Copper has no effect up to 0.5% on the toughness of submerged arc or inert gas shielded metal arc welds, as demonstrated in Figure 38, and also did not affect the hardness significantly. Copper additions over 1.0% increased strength but lowered the toughness (77) because of intergranular precipitation of copper. Other work (48) shows a slight decrease in toughness in CO_2 shielded MIG welds from the presence of 0.4% Cu.

Aluminum up to 0.4%, as indicated in Figure 37, improves the toughness of shielded arc welds. Titanium, also a deoxidizer and grain refiner, increases the toughness of MIG melts in small amounts from 0.008 to 0.01% (75).

Both sulfur and phosphorus decrease the toughness of welds. The influence of phosphorus and sulfur on weld metal deposited from basic electrodes has been studied (78). The volume fraction of inclusions increased linearly with higher sulfur contents resulting in lower upper shelf or ductile fracture energies in the Charpy test and reducing the COD value at which fibrous tearing first occurred. Phosphorus also reduced upper shelf Charpy V-notch fracture energies (78) and the toughness of the HAZ, as demonstrated in Figure 39.

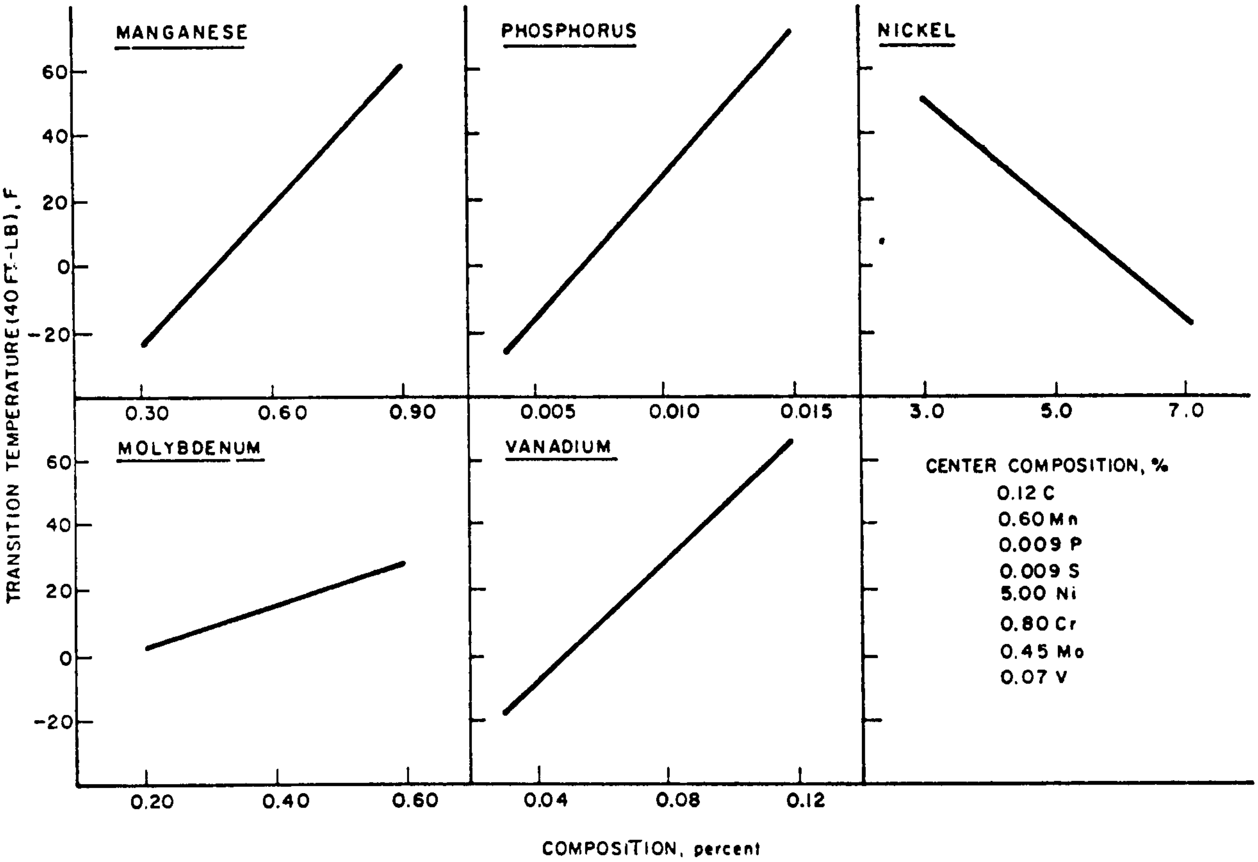


Figure 39. Predicted effects of composition on the stress relieved Heat-Affected Zone toughness of the center-composition steel shown at lower right (74).

Oxygen and nitrogen decrease the notch toughness of weld metals(48). The oxygen and nitrogen contents that can be expected from different welding processes are illustrated in Figure 40. Based on these results, the toughness of inert gas shielded welds would be expected to be better than submerged arc welds and those, in turn, superior to the shielded metal arc welds. This behavior is illustrated by the curves in Figure 27.

c. Microstructures -- Weld metals, with the possible exception of the electroslag process, are cast microstructures that have been solidified rapidly under a high thermal gradient. These conditions produce a cellular-columnar or columnar grain structurec In multipass welding, the original dendrite size and spacing is finer than in single weld deposits of the same size; in addition, the subsequent weld deposits recrystallize the columnar grains of the previous welds to produce finer equiaxed grains. These finer equiaxed grains are known to produce better Charpy V-notch toughness than coarser grains. A relation between the grain diameter and the 15 ft.lb Charpy V-notch transition temperature has been established for shielded metal arc welds (48):

$$T = A-B \log e \, (d^{-1/2})$$

where T is the transition temperature, A and B are constants, and d is the mean grain diameter.

This relation may predict somewhat lower values of transition temperatures in some cases where the structure contains excessive amounts of unrefined structures (48). The yield strength of the deposited welds increases with decreasing grain size.

d. Heat Inputs -- As the heat input increases, the rate of solidification and cooling in the weld metal and rate of cooling in the HAZ decreases. This difference in rate of cooling affects the toughness of the welds significantly. A comparison of the effects of heat inputs of 1.2 and 2.2 kilojoules per millimeter on the transition curves of the weld metal, using shielded metal arc electrodes of types E7016 and E7018G on rolled steel of 0.14% C and 1.07% Mn, are shown in Figure 41(79). The welds deposited in the flat positions have generally higher toughness than those welded in the vertical position. The higher heat input results in lower toughness for the 25 mm square COD tests, but the effect of heat input is not apparent in either the 10 mm COD tests or the Charpy V-notch transition Curves (79).

The influence of heat input on the toughness of submerged arc and MIG welds deposited in a groove in a low carbon 3/4 inch thick steel plate was also determined (73). The results of varying the heat input over a considerable range on the Charpy V-notch 15 ft.lb transition temperatures and strength levels are presented along with the chemical analysis in Figure 42 (73) for both submerged arc and (GMAW) weld deposits.

In the submerged arc deposited metals, the transition temperature increased to a maximum at a heat input of about 120 kilojoules per inch and then decreased slightly as the heat input increased to 231 kilojoules per inch (48). The transition temperature of the MIG welds increased

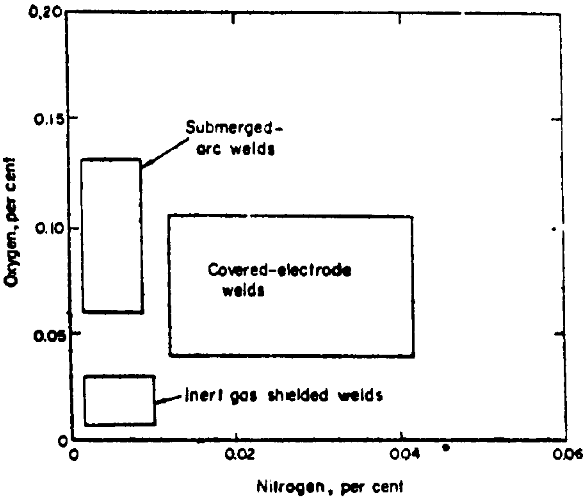


Figure 40. Oxygen and nitrogen contents of mild steel welds obtained in three arc welding processes (48).

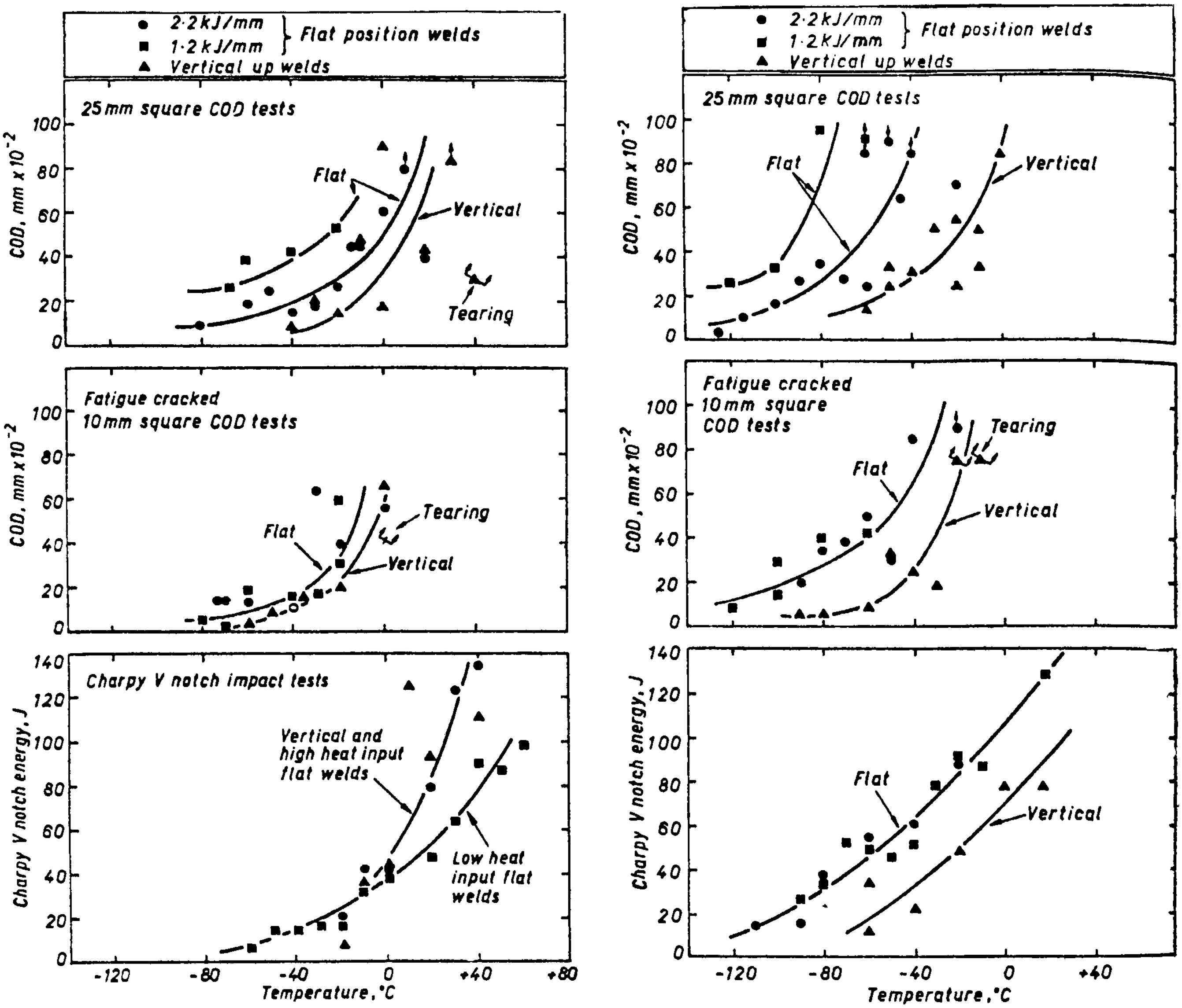
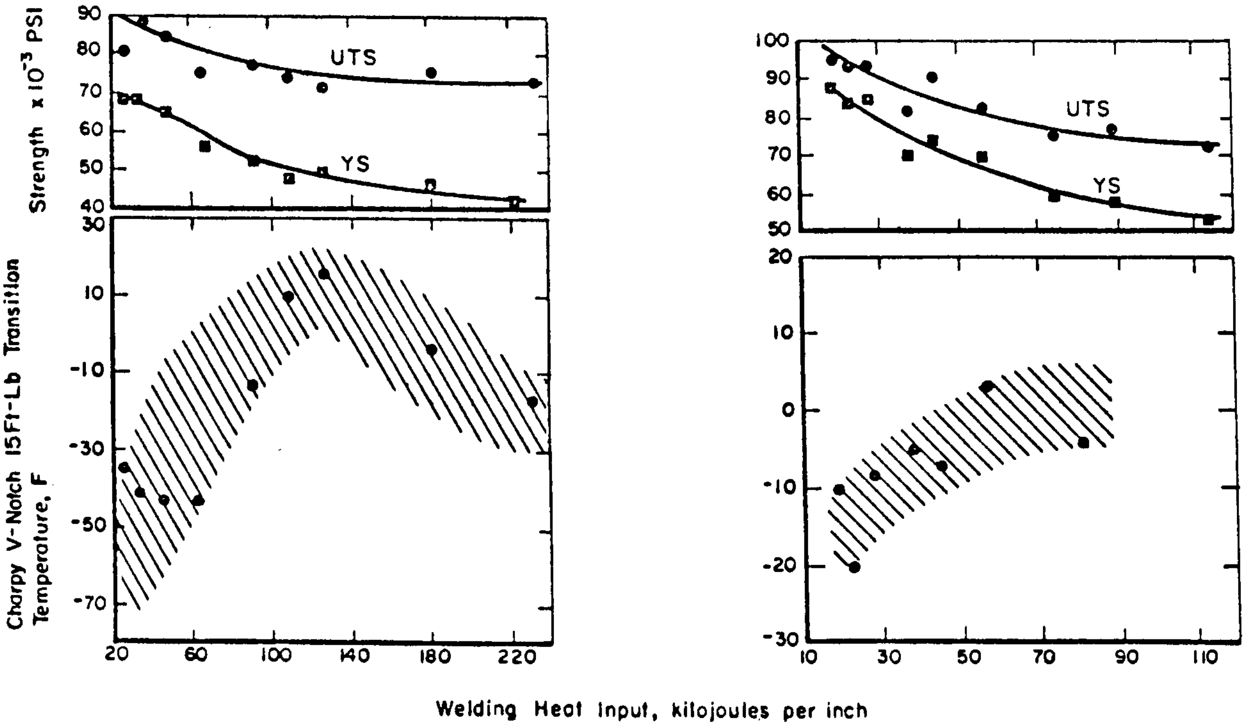


Figure 41. COD and Charpy transition curves on weld metal from E7016 electrodes (left) and E7018G electrodes (right) deposited in the flat and vertical position with two different heat inputs in the flat position (79).

Conversion:						
°C	-120	-80	-40	0	40	80
°F	-184	-112	-40		104	176



a. Submerged-Arc Welding

b. Inert-Gas Metal-Arc Welding

Figure 42. Effect of welding heat input on the notch toughness and strength of submerged arc and GMAW (MIG) weld metals (73).

	C	Mn	Si	P	S
Submerged-Arc Weld Metal (a)	0.09	1.40	0.34	0.014	0.023
Inert-Gas Metal-Arc Weld Metal(b)	0.08	0.90	0.37	0.011	0.018

steadily with larger heat inputs up to 81 kilojoules per inch or the maximum investigated. The strengths of both weld deposits decreases steadily, and the elongation and reduction of area increased with higher heat inputs.

The loss in toughness with higher heat inputs was attributed to the coarsening of the microstructure. A reduction in strain aging caused by weld restraint was attributed to the slight improvement in toughness of the submerged arc welds at the higher heat inputs. As previously discussed under submerged arc welding, the toughness also usually improves with a larger number of smaller weld passes because of the finer structure obtained.

The influence of heat input on toughness occurs primarily because of the effect that this factor exerts on the cooling rate. In a report on the development of low alloy, high strength thick plates with good weldability (80), it is shown that the toughness increases with a shorter cooling time between 1470 and 930°F for multilayered welds .

The effect of welding position on the toughness has been studied by several investigators with the general agreement that toughness decreased with out-of-position welding. The best toughness is obtained in the flat position, followed by horizontal (overhead) with the vertical position yielding the poorest toughness. This effect of position was shown to some extent in Figure 41. Other investigators (49) clearly demonstrate that the toughness obtained in the Charpy V-notch transition curve is significantly better in the flat position with toughness decreasing for the horizontal and vertical positions in that order. The lower toughness of the overhead and vertical welds results primarily from larger amounts of inclusions in these positions. Heat inputs can be higher in the flat position, and this can counteract the effect of position to some extent.

The effect of electrode size in the shielded metal arc process upon Charpy V-notch toughness of low hydrogen weld metal deposits has been investigated (49). Welds made with electrodes of a 1/4, 3/16, 5/32 inch diameter generally had lower Charpy V-notch values with the larger electrode diameters. This behavior was attributed to: the effect of changes in the electrode coating and its behavior on the composition of the weld metal; and the higher heat inputs that normally occur with the larger electrode diameters.

Dilution of the weld filler metal takes place when the parent plate and filler metal of different analyses are mixed during fusion. The influence of dilution depends on the relative composition of the weld deposit and base metal. Dilution of an E11018 electrode that was produced by using a thinner plate and narrower groove design has produced an increase in the Charpy V-notch 20 ft.lb transition temperature of approximately 60°F. In contrast, dilution which occured when welding a structural steel with an E7018 electrode decreased the fracture transition temperature of the weld deposit and increased the alloy content of the weld (49).

e. Heat Treatments -- Preheating reduces the thermal gradients in the weld and slows down the cooling rate of the weld deposit. Preheating has a definitely beneficial effect on the Charpy V-notch toughness of low carbon and low alloy steels. Increasing the preheating temperature of an E7018 weld deposit from 100 to 300°F reduced the Charpy V-notch transition temperature from +50 to -40°F (49). Increasing the preheat temperature from room temperature to 200, 400 and 600°F for E7015 and E8015-C2 electrodes steadily improved the Charpy V-notch toughness (81). In low alloy, high strength weld metals, the effect of preheat varies with the coating and weld metal analysis (48). It is reported (49) that an increase in preheat and interpass temperatures from 100 to 300°F had little effect on the notch toughness of weld metal deposited with an E9018 electrode.

Postweld heat treatments used in the fabrication of welded structures are classified as follows (48):

(1) Local heating employed shortly after welding.

(2) Stress-relieving heat treatment made at a temperature below the A_1 temperature.

(3) Normalizing and full annealing by heating above the A_1 temperature .

(4) Quench and tempering and other special heat treatments.

Stress relieving in the subcritical range of 1100 to 1250°F is the most commonly used postweld heat treatment, although higher temperatures are also employed. Most of the discussion in this section concerns stress relieving heat treatments.

The effect of stress relieving heat treatments (including some above the A_1 temperature) on the notch toughness of carbon and alloy steel weld metals has been investigated (48). Studies were made on electrodes of seven types: ilmenite (containing iron titanate), cellulose, iron oxide, titania, low hydrogen (Mn-Si steel wire), low hydrogen (Mn-Si-Cr-Mo steel wire) and iron powder. Butt welds 5/8 inch thick were heat treated for 1 hour at temperatures between 1110 and 1830°F and then furnace cooled. The yield strengths of the as-deposited weld metals ranged between 50 and 92 ksi. The 15 ft.lb transition temperatures obtained on Charpy V-notch tests of these weld metals are shown in Figure 43 (48).

The heat treatments caused losses in the notch toughness of weld metals made with electrodes other than the low hydrogen types, especially when the heating temperature exceeded the A_1 temperature. These stress relieving treatments produced little change in the notch toughness of weld metals made with low hydrogen-type electrodes; even slight improvement was observed in some cases. When the heating temperature exceeded the A_1 temperature, a marked increase in grain size was observed in the weld metals made with all types of electrodes (48).

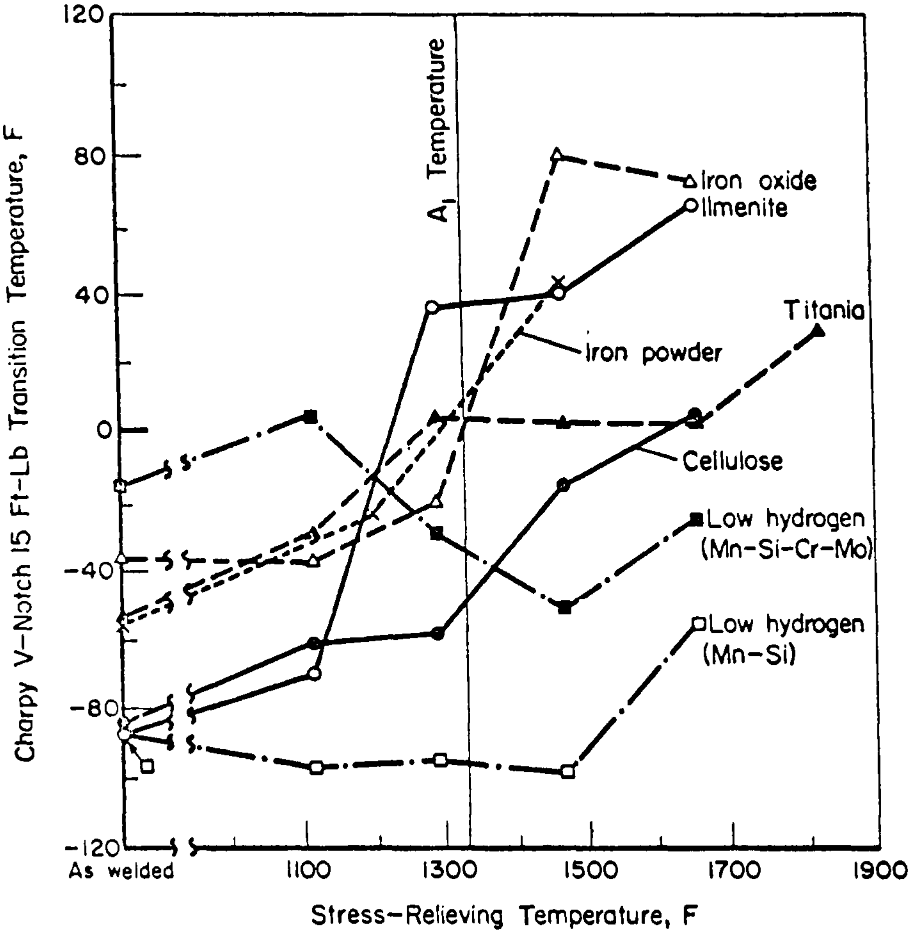


Figure 43. Effect of stress relieving on the notch toughness of weld metals deposited with various types of electrodes (48).

The stress relief embrittlement of both carbon and alloy steel weld metals has been studied (49). It is reported that weld metals which undergo secondary hardening (increasing the tensile strength) embrittle to the largest extent as a result of stress relief, whereas an improvement in notch toughness can be expected when the tensile strength decreases significantly after stress relief. Vanadium is known to promote secondary hardening and stress relief embrittlement. The Charpy V-notch transition temperature of a weld metal made with E10015 electrodes of Ni-Mo-0.1 V wire increased by about 40°F after stress relieving for 2 hours at 1150°F, while the notch toughness of a weld metal made with similar electrodes of vanadium free wire was barely affected by the stress relieving heat treatment (48).

Other investigators on the effect of stress relief have noted improvements in toughness under some conditions. The toughness exhibited by the Charpy V-notch test showed a slight increase with stress relieving compared to the as-welded condition for four of seven types of electrodes tested (E6013, E6020, E7015 and E8018-C2); the toughness of the other three was reduced somewhat more (E6010, E8015-C2 and E10016) (81). Other work (49) observed a significant improvement in the Charpy V-notch toughness of E7016 and E7018 electrodes as a result of an 1150°F stress relieving treatment. The effect of an lll0°F stress relieving treatment for one hour on several different weld metals and welding processes has been studied using fatigue cracked COD specimens (79). The influence of this stress relieving treatment on the COD test results are listed in Table XI (79). A loss of toughness occurred with stress relieving in all manual welds except two (an E7016 and E6018 deposit), and the improvement was minor in these cases. The toughness of the submerged arc deposit was unaffected; the flux-CO_2 shielded and bare-CO_2 shielded wire lost toughness with stress relieving. The COD properties of the electroslag deposits improved significantly. Other investigators used fracture toughness tests (87) and COD techniques (83) to measure the toughness of weld deposits after welding and stress relieving with about the same results as shown above.

Electroslag welding with the large grain size produced in the weld metal and HAZ by the high heat inputs can be significantly improved by normalizing treatments after welding. These improvements in Charpy V-notch toughness for both the weld metal and HAZ have been observed for both normalizing and double normalizing treatments after welding (84).

Different combinations of preheat and postheat have been investigated for single-pass weld metals of nominal analysis - 0.10% C, 1.4% Mn, 0.40% Si, 0.20% Ti. An arbitrarily chosen displacement of 0.l0 mm was designated as the COD fracture transition, Cycling the test pieces to 1830°F was employed to simulate a later weld pass. Table XII (85) and Figure 44 (85) summarize the results.

Three distinct areas of toughness were observed: a high toughness was obtained with the simulated later weld (K and L); an intermediate toughness (A, B, C, D) for welds which had no preheat or postheat; and a low toughness (E, F, G. H, J) for welds which had undergone a 930°F preheat, 1155°F postheat or both.

TABLE XI: TOUGHNESS TESTS ON 0.4 IN. (10mm) SQUARE FATIGUE CRACKED COD SPECIMENS IN THE AS-WELDED AND THERMALLY STRESS RELIEVED CONDITIONS (79).

		COD,mm x 10^2	
Welding Electrode and Process	Test Temperature °F	As Welded	Stress Relieved 1110°C for 1 Hour
E7013 (SMAW)	+14	>65	6.0
E7016 "	-4	37	10.0
	-40	18	2.5
E1016G "	+32	25*	38.0
	-40	37	1.0
	-76	24	3.5
E7018 "	-40	61	3.5
	-112	3	1.0
E7018G "	-4	>50	3.5
	-76	43	4.0
	0	36*	4.5
E9018G "	-4	50*	3.0
	-76	25	1.0
E1016G "	-4	38*	10.5
	-76	27	1.0
	-4	41*	30.0
E8016 "	-40	40*	10.0
	-112	21	5.5
	0	41*	57.0*
E7016 "	-4	59,31	64.0
	-40	16	3.5
	-112	5	3.5
E8016-C1 "	-4	>65	6.0
	-76	>65	7.5
E6013 "	-4	>65	5.5
	-76	34.5	2.0
	-40	65	5.0
	-76	40.5	2.5
	+59	60	65.0
	0	66	25.5
	-4	17.5	4.8
	-40	9	5.0
E6018 "	-148	45,21	45.0
	184	8	24.0
Electroslag	0	63,50	>65.0
	-4	21	3.5
	-40	17	24.0
Consumable guide	+104	19	36.0
Electroslag	+61	8.5	17.0
Submerged arc	+32	58	>65.0
	-4	11.5	15.0
	-40	12	3.0
Bare wire CO_2	-40	>65	3.0
Flux cored CO_2	-40	43	3.0
	-112	30	1.0

*Tearing

TABLE XII: PREHEAT AND POSTWELD HEAT TREATMENT CONDITIONS WITH 0.10mm COD TESTING TEMPERATURES (85).

Specimen	Preheat	Postweld Heat Treatment	0.10mm COD Temp. °F
A	None	None	-220
B	"	"	-184
C	"	"	-224
D	"	"	-181
F	"	1155°F/1 hour, furnace cooled	-132
G	"	1155°F/1 hour, water quenched	-153
K	"	Cycled to 1830°F	-274
E	930°F	None	-126
H	"	1155°F/1 hour, furnace cooled	-134
J	"	1155°F/1 hour, water quenched	-126
L	"	Cycled to 1830°F	-247

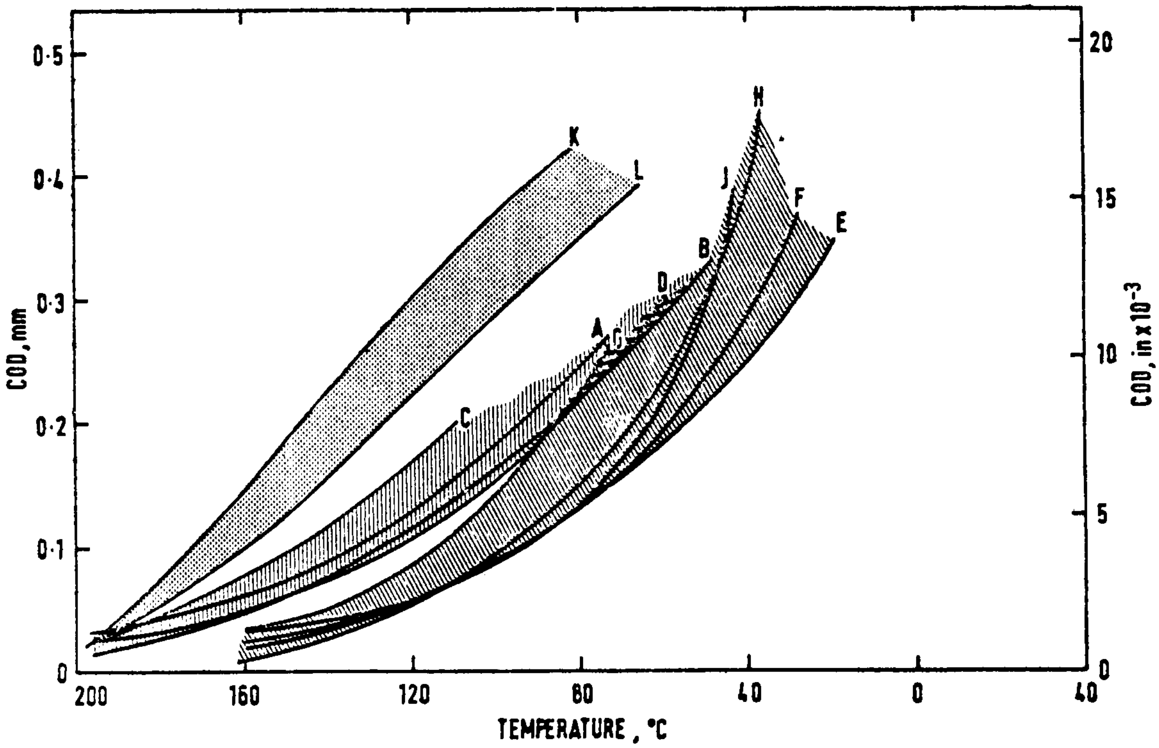


Figure 44. Results of COD tests on welds with different preheat and postheat treatments (85).

Conversion:						
°C	-200	-160	-120	-80	0	40
°F	-328	-256	-184	-112	32	104

In an investigation involving the arc welding of 5% Cr, 0.5% Mo steel pipe, the toughness of the weld metal was improved considerably by postheating at 1250 to 1400°F and improved further with a 500°F preheat and 350°F postheat (86).

Welding Discontinuities --

In an investigation sponsored by the Steel Foundry Research Foundation, the influence of weld discontinuities in cast 8630 steel on the impact resistance was measured (87). Tests were made by tension and bending impact techniques upon welded steel in either the normalized and tempered (N & T) or quenched and tempered (Q & T) conditions. The welding operation performed and the types of discontinuities are similar to those reported in reference 28 and shown in Figure 14 and Table IV, except that the TIG process with a filler rod similar to the E9018 composition was used for the tension impact bars. The bending specimens were tested as a simple beam in an oversize Charpy Type Machine that had a capacity of 2200 ft.lbs. The bend specimens were 7-1/2 inches long by 1 inch wide by 1/2 inch thick and tested unnotched. The tension impact specimens had a 0.357 inch gage diameter, 1.425 inches long with larger diameter ends for holding during testing. The steels were heat treated to the properties shown in Table XIII(87). The specimens were all welded as cast and completely heat treated after welding.

The results of these tests indicated that the loss in toughness from discontinuities was severe. The results show that the effect of these discontinuities, in order of decreasing severity for the bending impact was: undercut, surface slag, slag inclusions located internally in the specimen and incomplete penetration. The tension impact tests demonstrated the greatest loss of toughness from slag inclusions followed by undercut and incomplete penetration. The sound, welded specimen with the weld reinforcement in place exhibited a loss in toughness as indicated by reduced fracture energies because of the notch effect at the edge of the reinforcing bead. The toughness of welded cast steel with the weld machined to the specimen contour was similar to cast steels process to the same strength level by a similar heat treatment.

The effect of the various types of welding discontinuities in 8630 cast steel specimens on the fracture energy levels and temperatures for the start of ductility, 50% ductility and 90-100% ductile behavior are listed in Table XIV (87).

Weld undercut affected weld toughness more in bending than in tension impact testing because of its location at the surface. Impact energies above the completely fibrous transition temperature in bending were decreased by 60 to 65% by undercuts as compared to sound welds for Q & T and N & T welds, respectively; transition temperatures increased 125°F for Q & T samples and 75°F for N & T samples. The notch effect of undercuts in tension was not as great as procuced by slag inclusions in tension. Slag inclusions in welded Q & T specimens reduced impact energies about 65% compared to sound welds in both

TABLE XIII: MECHANICAL PROPERTY DATA FOR WELDED STEEL IMPACT SPECIMENS TESTED FOR EFFECT OF DISCONTINUITIES (87).

Heat Treatment	Tensile Strength 1000 psi	0.2% Y. S. 1000 psi	Red. of Area %	Elong. in 1.4 In. %	Brinell Hardness Number	Impact Specimens
Quench & Temper	149.3	141.1	25.6	13.2	319	Tension
Normalized & Temper	93.8	59.6	31.9	23.1	153	Tension
Quench & Temper	134.1	119.9	33.6	14.8	264	Bending
Normalized & Temper	90.4	62.6	33.4	21.7	187	Bending

TABLE XIV: IMPACT TRANSITION DATA OF SELECTED CRITERIA DEMONSTRATING THE INFLUENCE OF WELDING DISCONTINUITIES (87).

Condition	Type of Test	Discontinuity	First Indication of Ductility		50% Fibrosity Transition		90-100% Fibrosity Transition	
			(Ft.-Lbs.)	Temp. (°F)	(Ft.-Lbs.)	Temp. (°F)	(Ft.-Lbs.)	Temp. (°F)
Q&T	Bending	Sound as	79	∼−321	60	∼−290	1220–1500	−200
Q&T	Tension	Machined	80	−250	80	−250	300	−200
N&T	Bending	Sound as	106	−200	730	−150	1300	−100
N&T	Tension	Machined	126	−200	125	−200	380	−110
Q&T	Bending	Sound as Welded	—	<−250	400	−290	950–1200	−200
Q&T	Tension	Sound as Welded	∼60	−250	60	−250	260	−175
N&T	Bending	Sound as Welded	120	−200	550	−130	1250	−100
N&T	Tension	Sound as Welded	75	−200	110	−175	280	−100
Q&T	Bending	Slag Porosity	70	−200	210	−125	560	−75
Q&T	Tension	Slag Porosity	40	−150	60	−110	95	−50
N&T	Bending	Slag Porosity	100	−200	500	−50	730	−25
N&T	Tension	Slag Porosity	40	−110	90	−25	150	0
Q&T	Bending	Undercut	35	−200	150	−125	500	−75
Q&T	Tension	Undercut	80	−200	120	−175	200	−150
N&T	Bending	Undercut	230	−125	350	−75	550	−25
N&T	Tension	Undercut	70	−150	135	−125	230	−110
Q&T	Bending	Incomplete Penetration	100	−250	250	−200	630	−150
Q&T	Tension	Incomplete Penetration	50	−200	100	−175	220	−150
N&T	Bending	Incomplete Penetration	220	<−150	470	−125	800	−60
N&T	Tension	Incomplete Penetration	50	−150	100	−125	250	−110

bending and tension impact tests; the transition temperatures increased 125° F for bending and 150°F for tension. In welded N & T samples, slag inclusions decreased impact energies about half for both types of tests and increased transition temperatures by approximately 100°F. Incomplete penetration reduced fracture energies approximately 50% for the quenched and tempered welds and about 40% for normalized and tempered welds and increased transition temperatures about 50°F in both types of tests. Incomplete penetration reduced fracture energies of tension impact specimens about 25% and increased fracture transition temperatures by 50°F in Q & T samples; no shift in transition temperatures occurred for N & T specimens (87).

Another investigation of the effect of discontinuities on the toughness of welds has shown that the presence of microcracks reduces the toughness of Charpy V-notch specimens, as measured by the ft.lbs for fracture, by about one-half over a testing temperature range of room temperature to -100°F for E6010 weld deposits (88). Other studies discuss the influence of slag inclusions in promoting discontinuous crack propagation through a steel weld (89) and the influence of hot straining (46, 90).

IV. APPLICATION OF FRACTURE TOUGHNESS CONCEPTS TO MECHANICAL BEHAVIOR OF WELDS

The fracture toughness concepts and tests discussed in the section on types of tests for toughness offer the possibility of calculating the maximum allowable or tolerable crack sizes that can be permitted in weld deposits without failure occurring. Tests that indicate the transition from ductile to brittle fracture, such as the Charpy V-notch, drop weight and dynamic tear tests, cannot be readily utilized for this purpose. However, the K_{Ic} value and COD (Crack Opening Displacement) tests can be employed directly (91). The K_{Ic} value is employed for these crack size calculations when fracture occurs with only a small amount of plasticity at the crack tip. When significant amounts of plasticity precedes fracture at this crack tip, such as frequently occurs in commercial welds made in medium and light sections, the COD test results may be employed for allowable crack size determinations (91). The calculation of these crack sizes and their effect on failure of the weldment provides a means of determining the acceptable limit of weld discontinuities in non-destructive testing.

A similar approach of using fracture mechanics to determine the acceptable limits of discontinuities in welds subjected to fatigue stresses also offers promise (92). Further work is required to obtain additional data on the behavior of various welded configurations and materials so that the method can be generally applicable. When these data are available on various materials and weldment shapes, the design computations possibly will be much more adjustable to various service conditions than from simple S-N curves (92).

V. SUMMARY AND CONCLUSIONS

The weldability of cast steels is broadly defined as their ability to provide a quality weld joint with good mechanical properties and without the need for special welding process precautions or treatments. The types and causes of mechanical and metallurgical welding discontinuities that may occur and the effect of various welding processes and techniques on their occurrence are described.

Under this weldability definition, low carbon cast steels (C ≥0.20% C) have excellent weldability and only limited problems with gas porosity and hot cracking. These difficulties can be readily prevented by control of the deoxidizers and sulfur content of the steel. Medium carbon steels with a carbon content in the 0.20 to 0.50% range have good weldability up to 0.30% C but become increasingly difficult to weld at higher carbon contents requiring preheating temperatures up to 400°F under some conditions. Welding with low hydrogen electrodes or with processes employing a low level of hydrogen in the shielding atmosphere reduce or eliminate the need for preheating. Stress relieving after welding is recommended over 0.40% C. High carbon cast steels (0.5-1.0% C) are difficult to weld because of the possible formation of brittle high carbon martensite in the heat affected zone. With low hydrogen welding atmospheres, preheating temperatures as high as 600°F and postheating by stress relieving or preferably a full heat treatment, satisfactory welds can be produced. Thicker section welds require greater precautions.

The alloy steel grades are discussed in this paper up to 5% alloy contents. Their weldability varies widely from low carbon types that contain considerable alloy for strengthening to the medium alloy 4300 types. The low carbon, alloy steels have good weldability with several processes and only require low preheating temperatures (100-200°F) in heavy sections. As the carbon and alloy contents become higher, the hardness levels attained in the heat affected zone increase with a greater tendency for cold crack formation. Prevention of this cracking requires preheating to higher temperatures, low hydrogen welding atmospheres, and lower sulfur and phosphorus contents. The higher alloy types are generally fully heat treated after welding.

The fatigue properties of weld joints are frequently significantly inferior to the base casting. This loss in fatigue properties arises from a number of causes and can be minimized by selective welding procedures. While the weld fatigue strength increases somewhat with the strength of the base steel and weld deposit, this increase levels off at about a base casting strength of 105 ksi with an endurance limit in the range of 40 to 50 ksi. The fatigue limit of welds is below that attainable with higher strength cast steels even when the weld reinforcement is ground flush to the casting surface and with a full heat treatment.

Welds that are made with low hydrogen shielding atmospheres and automatic welding processes (such as submerged arc and electroslag welding) have better fatigue properties compared to manual processes primarily because of the reduction in welding discontinuities. Relatively small discontinuities can cause significant losses in fatigue properties. The magnitude of the effect of such discontinuities depends on the fatigue stressing conditions; losses in endurance ratio of up to 50% can result from surface and internal porosity, undercuts, slag inclusions and lack of penetration.

Postwelding treatments may improve the fatigue properties of weldments. A full heat treatment is very effective if stress concentrations from weld reinforcements and discontinuities are avoided. Stress relieving improves the fatigue strength under alternating compression and tension loads but not for purely tension stresses. Peening of the as-welded structures and grinding the weld flush to the weld surface are both effective in improving fatigue behavior.

The toughness of welds has been primarily measured by Charpy V-notch specimens utilizing a number of transition temperature, minimum fracture energy and fracture appearance criteria. Recently, fracture toughness methods such as the plane strain fracture toughness parameter, K_{Ic}, and the crack opening displacement, COD, have been employed. These tests have shown that the notch toughness of welds is markedly affected by a large number of welding and postwelding processes in addition to the influence of microstructure, composition and strength level that have been established as significant factors for the toughness of cast steel.

The toughness of weld metal in steel castings is improved by the following welding procedures:

(a) use of Gas Tungsten Arc (GTAW or TIG) , Gas Metal Arc (GMAW or MIG) and submerged automatic arc welds compared to manual deposited, shielded arc welds;

(b) low hydrogen welding electrodes or atmospheres with basic fluxes;

(c) reduced heat input during welding and multiple welding passes compared to single or a few larger weld deposits;

(d) a low carbon, oxygen, nitrogen, sulfur and phosphorus content in the weld deposit with no more alloy than required for strength; nickel is less damaging and vanadium more detrimental than other alloys; aluminum deoxidation is generally beneficial.

(e) a flat welding position, followed by horizontal (overhead), with the vertical position the poorest;

(f) reduced welding restraint.

The notch toughness of the heat affected zone was lowest in the grain coarsened portion, better in the grain refined section and best in the intercritical region. Postweld heat treatments improve the toughness in the heat affected zone and either decrease or increase the weld metal toughness depending on the microstructure and composition. A full heat treatment provides better toughness in the heat affected zone than stress relieving. The high heat input of electroslag welds results in a coarse grain size in both the weld deposit and the heat affected zone with low toughness, requiring a full heat treatment for some improvement.Increased levels of alloy reduce the toughness of the heat affected zone in as-welded steels.

The presence of discontinuities can sharply reduce the notch toughness of welds in cast steel. The general effect of these discontinuities listed in order of decreasing severity is cracks, undercuts, slag inclusion and lack of penetration, although the location of the discontinuity with respect to the maximum stress distribution can affect this order.

The fracture mechanics approach and the use of available fracture toughness data allows calculations to be made to determine the effect of various discontinuities on the toughness and fatigue behavior of welds. These calculations can be employed to provide maximum safe limits of discontinuities measured by non-destructive testing.

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