1-1-1982

Hydrogen induced cracking in flux cored electrode welds.

Srinivasarao Lathabai

Recommended Citation

Lathabai, Srinivasarao, "Hydrogen induced cracking in flux cored electrode welds." (1982). *Theses and Dissertations.* Paper 2008.

HYDROGEN INDUCED CRACKING IN FLUX CORED ELECTRODE WELDS

by

Srinivasarao Lathabai

A Thesis
Presented to the Graduate Committee
of Lehigh University
in Candidacy for the Degree of
Master of Science
in
Metallurgy and Materials Engineering

Lehigh University
1982

CERTIFICATE OF APPROVAL

This thesis is accepted in partial fulfillment of the requirements for the degree of Master of Science.

9·3-82
(date)

Professor in Charge

Chairman of the Department

ACKNOWLEDGEMENT

The author acknowledges a deep sense of gratitude to Dr. R. D. Stout for his guidance and support during the course of this work. She is grateful to Dr. A. W. Pense for his encouragement and interest. Thanks are due to Messers. D. Conklin, J. Williams and W. Mohylsky for being so prompt in machining test specimens and to Betty Zdinak for typing the manuscript.

Finally, the author acknowledges the financial support of this research program by the American Iron and Steel Institute and Lehigh University for its support through the Distinguished Fellowship.

TABLE OF CONTENTS

LIST OF TABLES

LIST OF FIGURES

ABSTRACT

In recent years, flux cored arc welding (FCAW) is being increasingly used in fabrication of heavy weldments of mild and low alloy steels. The process combines the advantages of a continuous process with the metallurgical benefits resulting from using a flux. Its high productivity makes it competitive to the shielded metal arc welding process in a variety of applications.

While tremendous progress has been made in the development of flux cored wires, there is little published research on the quality of the weld metals deposited by this process. FCAW is generally assumed to be a low-hydrogen process, but actual values of diffusible hydrogen contents in different types of flux cored wires are not readily available.

An investigation was undertaken to determine what hydrogen levels could be encountered in a representative series of flux cored wires, covering a range of strengths and flux types. The susceptibility of welds made using these flux cored wires on a variety of steels to hydrogen induced cold cracking was evaluated. The diffusible hydrogen contents of the different flux cored wires were evaluated by means of hydrogen evolution tests using glycerin as the collecting medium. The tests, conducted as per the recommendations of the Japanese Standard, JIS Z 3113-1975, revealed that diffusible hydrogen contents of the rutile type electrodes can range between 5-13 ml/100 gms of

deposited metal while the basic flux cored wires produced 1-1.5 ml/100 gms of deposited metal.

The hydrogen induced cracking susceptibility of different steels was assessed by the modified implant test. The response of the steels varied with their hardenability. The flux cored wires induced sensitivity to hydrogen assisted cracking in hardenable steels such as ASTM A302B and A517F, the sensitivity in general being proportional to the diffusible hydrogen contents of the wires. Increasing the heat input had no significant effect on lowering the sensivitity. On a low hardenable structural steel, A588B, some cracking tendency was observed at a low heat input level, but doubling the heat input removed this sensitivity. Flux cored wires induced cold cracking in welds made in A36 steel but no sensitivity was observed in the pipeline steel X65. Results of Lehigh restraint tests, carried out on some of the filler metal-steel combinations, were in agreement with the implant test results.

From the results of the investigation, it may be concluded that flux cored wires can engender fairly high levels of diffusible hydrogen in weldments. So, in the application of flux cored wires to structural and pressure vessel steels, preheating may be needed for the higher carbon steels and the high strength low alloy steels, particularly in sections over one inch thick.

INTRODUCTION

The flux cored arc welding process was first introduced in the United States in 1957.(1) Since then, tremendous progress has been made in the flux cored wire industry, until today, when the process offers competition to the shielded metal arc welding and gas metal arc welding processes in terms of higher productivity and economic efficiency.

Flux cored arc welding (FCAW) is an arc welding process which produces coalescence of metals by heating them with an arc between a continuous filler metal (consumable) electrode and the work.(2) Shielding is provided by a flux contained within the tubular electrode and may be supplemented by an externally applied gas. FCAW offers two major process variations that differ in the method used to shield the arc and the weld pool from atmospheric contamination. One type, self-shielded FCAW, protects the molten metal to some extent through the decomposition and vaporization of the flux core by the arc (Figure 1). The other type, gas shielded FCAW makes use of a protective gas flow in addition to the flux core action to shield the arc and the weld pool (Figure 2). With both methods, the electrode core material provides a relatively thin slag covering to protect the solidifying weld metal.

The main reason for the increasing use of flux cored arc welding in steel fabrication is that it combines the productivity of continuous welding with the metallurgical benefits that can be derived from a flux. Thus the FCAW possesses some of the characteristics of the shielded metal arc welding (SMAW), the gas metal arc welding (GMAW) and the submerged arc welding (SAW). Consequently, in certain applications, each of the two FCAW process types offers certain trade off benefits over the related processes. This is especially true in the case of SMAW. In the ship building industry, for example, the recent trend has been replacement of SMAW with FCAW to increase productivity and reduce welding time substantially.(3)

The growth of the FCAW filler metal industry has been vigorous. In recent years, flux cored electrodes have been developed for all-position production welding and these are being constantly improved in their quality and performance.(4) The so-called "second generation" electrodes are now being marketed, which combine excellent operational characteristics with weld metals of good quality.(5) This continuous growth in the filler metal industry has provided the impetus for the increasing use of FCAW in all types of heavy fabrication involving mild and low alloy steels. FCAW has been used in the manufacture of railroad hopper cars, bearing supports, box girders, the main frames for heavy duty hydraulic presses,(6) power shovels,(7) tapered steel poles for high voltage transmission lines,(8) and pipe welding in nuclear plants.(9)

While there has been intense research going on in the filler metals industry to develop new and better flux cored wires, there is very little published work on the assessment of flux cored weld metal mechanical properties, especially toughness and the susceptibility of the weldments to different forms of cracking. Flux cored wires have been assumed to exhibit a low content of hydrogen and, therefore, little susceptibility to hydrogen induced cracking. As with covered electrodes formulated with low hydrogen materials, however, the possibility exists that flux cored electrodes may have moisture or other hydrogen containing materials entrapped during manufacture or will pick up hydrogen, usually as moisture, between their manufacture and use. The necessity of storing low hydrogen covered electrodes in an oven until shortly before use is well recognized, but similar precautions are not always taken for flux cored electrodes. The AWS D1.1 code states only that they should be "dry and in suitable condition". The Guide for Specifications for Fracture Critical Non-redundant Steel Bridge Members, published by AASHTO, does state that the electrode manufacturer should provide the fabricator with a written description of the storage requirements that must be followed to keep the hydrogen content of the materials low enough to meet certain limits for the concentration of diffusible hydrogen in the deposited metal.(12)

On account of the above facts, the Weldability Committee of the Welding Research Council sponsored an investigation at Lehigh

University. The main objectives of this study were to determine: (i) the extent to which the various types of flux cored electrodes used in steel construction introduce hydrogen into welds and (ii) the response of low carbon structural steel and high strength low alloy steel weldments to delayed cracking when produced with flux cored electrodes. This thesis is a presentation of the results of this investigation.

BACKGROUND

Hydrogen Induced Cracking in Weldments

Hydrogen induced cracking, also referred to in the literature as cold cracking, delayed cracking, underbead cracking and internal hydrogen embrittlement, is in general one of the most important potential sources of defects in welding hardenable steels. Cracks in structures cannot be tolerated to any level, since they can lead to catastrophic brittle failures in service, resulting in loss of life and property.

While the damaging effects of hydrogen in steels have been recognized and researched for more than a century,(13-16) a complete understanding of its mechanism has remained elusive. This is particularly true in the case of hydrogen-induced cracking in weldments. However, the conditions under which hydrogen induced cracking occurs in steel have been clearly established.(14) Three factors have to be present simultaneously for cracking to occur: (i) a source of hydrogen which provides a concentration of it in the weld metal; this could be moisture, hydrogen bearing compounds in the electrode or contamination, such as grease, in the weld joint, (ii) a unfavorable, brittle, low temperature transformation product in the heat affected zone (HAZ) of the weld (this depends on the cooling rate and chemistry of the steel), and (iii) restraint and reaction stresses in the weld joint.

Hydrogen embrittlement, in general, refers to the severe degradation of physical and mechanical properties by the action of hydrogen. Depending upon the manner in which the hydrogen-metal interactions occur, hydrogen embrittlement has been classified into three broad categories:(15) (i) hydrogen reaction embrittlement, (ii) hydrogen environment embrittlement, and (iii) internal hydrogen embrittlement (also called hydrogen-assisted cracking).

Hydrogen reaction embrittlement is based on the reaction of hydrogen with itself, with the matrix or with an alloying element in the matrix. The source of hydrogen can be the atmosphere during melting, pickling, electroplating, etc. Once the hydrogen is absorbed, it may react near the surface or diffuse substantial distances before it reacts. The hydrides formed by these reactions are found to be quite stable and the embrittlement is not reversible at room temperature. The atomic hydrogen can also react with itself to form molecular hydrogen at discontinuities, resulting in high pressures to form defects. Yet another possibility is that the atomic hydrogen can react with an alloying element in the steel to form a gas. An example is the degradation of mechanical properties of carbon steels caused by the high temperature reaction of carbon with hydrogen to form methane occlusions.(16) This type of hydrogen embrittlement is not considered responsible for the delayed cracking under study in this work.

The embrittlement of metals due to their exposure to hydrogen bearing atmosphere is called the hydrogen environment embrittlement. This embrittlement could be due to exposure to gaseous hydrogen atmosphere or due to interaction with an aqueous hydrogen-producing environment. Once the hydrogen has been absorbed into the metal lattice, the mechanism that has been postulated to explain this phenomenon is very similar to that proposed for Internal Hydrogen Embrittlement.(17)

It is this third category of hydrogen embrittlement, namely Internal Hydrogen Embrittlement, that is considered to be responsible for the delayed cracking in weldments. Hydrogen Assisted Cracking is another term that has been used to describe this phenomenon which results in a drastic reduction of toughness of the material. The hydrogen is charged into the material by several industrial processes such as electroplating, melting, pickling, welding, etc. No chemical reactions take place as in hydrogen reaction embrittlement. The other major factor that distinguishes this type of embrittlement is the fact that, unlike environement embrittlement where there is an unexhaustible supply of hydrogen, the phenomenon involves a fixed quantity of hydrogen being picked up during the process of welding.

As mentioned earlier, the hydrogen assisted cracking in weldments requires the presence of hydrogen, a susceptible microstructure and stresses. The following is a review of significant work done regarding the importance of these three factors, followed by a summary of the various mechanisms that have been proposed to exaplain the phenomenon of hydrogen assisted cracking.

A. Effect of Hydrogen

The fact that hydrogen dissolves atomically was established by Sievert.(17) The solubility of hydrogen at equilibrium and atmosphere pressure, in iron at room temperature is only about 20% of that in austenite before it transforms at 910°C.(18) The supersaturation of hydrogen in the ferritic matrix was attributed to be responsible for cracking. Since this discovery in 1940, several investigators using different types of tests, have established the effect of hydrogen in cracking. For example, Suzuki and co-workers, using the TRC test, showed that the critical stress for cracking decreased as the diffusible hydrogen content of the weld increases.(19) Evans and Christensen, using the implant test,(20,21) found a similar inverse relationship between hydrogen content and the static fatigue limit, σ_{SFL}. Beachum, Johnson and Stout(22) have reported that as little as 2 ppm of total hydrogen was sufficient to cause delayed cracking. It was found in this work that the time for cracking increased at lower aging temperatures, indicating the role of diffusion rate in the phenomenon. Interrante and Stout using the Lehigh Restraint test, showed that the susceptibility to delayed cracking was directly proportional to the hydrogen content in the welding atmosphere.(23) They also found that the use of an austenitic stainless filler metal resulted in eliminating delayed cracking, presumably by serving as a "reservoir" for hydrogen.

A major limitation in the study of hydrogen induced cracking is the problem of quantifying the critical concentration of hydrogen in the vicinity of damage. Hydrogen generally may exist in different locations in a metal, such as voids, dislocations, etc. which can serve as traps. It is the mobile, diffusible hydrogen that causes damage. Indeed, a beneficial effect of sulfides, which act as hydrogen sinks, in lowering the susceptibility to hydrogen cracking, has been reported.(24,25) The critical concentration of hydrogen has been assumed, in several investigations,(20,21) to be proportional to the diffusible hydrogen content of the weld. However, it has to be borne in mind that the critical hydrogen concentration is also a function of the microstructure and the stress level.

B. Effect of Microstructure

The fact that the susceptibility to hydrogen induced cracking is strongly dependent on the composition of the steel has been recognized for a long time. This effect of composition is sometimes expressed by the concept of carbon equivalence. The carbon equivalent in fact, is a measure of the hardenability of the steel and is similar to the statistically derived formulae for M_s temperatures. This is based on the observation that sensitivity to hydrogen cracking increases progressively as the M_s temperature becomes lower. As mentioned before, the presence of a low temperature transformation product in the heat affected zone (HAZ) is an essential condition for the hydrogen induced cold cracking to occur; and the higher the carbon equivalent, the greater is the tendency to cracking.

Of the alloying elements in steel, carbon is the critical one, since it controls the nature and properties of the low temperature transformation products, apart from contributing most to the hardenability of the steel. The other elements contribute to cracking sensitivity, mainly by increasing the hardenability. The most common formula to evaluate the carbon equivalent for carbon-manganese steels, is as follows:(26)

$$CE = C + \frac{Mn}{6} + \frac{(Cr + Mo + V)}{5} + \frac{(Ni + Cu)}{20}$$

When carbon equivalent formulae are used, the assumption is that below a certain value of CE, the danger of cracking is negligible. This is related to the type of microstructure that may be expected in the HAZ for a certain value of CE. Although precise information on the microstructural constituents in the HAZ of different steels is not available, the following assumption has been tentatively made about the microstructures of the HAZ for arc energy inputs as might be encountered in SMAW:(21)

1. Essentially, non-martensitic microstructures at low carbon equivalents ($CE < 0.3$).

2. Mixed structures in a transition range (CE from 0.3 to 0.35).

3. Essentially martensitic microstructures at CE > 0.35.

The critical value for CE would, however, vary with the hydrogen potential of the welding process being used. Thus, the CE value below which cracking will not occur would be substantially lower for cellulose electrodes when compared with that for low hydrogen electrodes.

Japanese investigators Ito and Bessyo have proposed a much more elaborate formula, taking into account restraint stresses and hydrogen.(27) They proposed a cracking parameter for welding high tensile low alloy steels as follows:

$$P_W = P_{CM} + \frac{H}{60} + \frac{K}{4000}$$

$$P_{CM} = C + \frac{Si}{30} + \frac{Mn + Cu + Cr}{20} + \frac{Ni}{60} + \frac{Mo}{15} + \frac{V}{10} + 5B$$

where H is the diffusible hydrogen content in ml/100 gms and K is the restraint intensity factor in N/mm^2 given by:

$$K = \frac{\text{Young's modulus x plate thickness}}{\text{Restraining length*}}$$

(*From TRC tests)

Winterton has discussed the various formulae proposed to date along with the composition limits, weld tests and the criteria on which these formulae were based.(28) The main drawback of the carbon-equivalent formulae is that they can only be applied to the type of steels for which they were experimentally derived. Thus, the various formulae could be conservative when applied to predict the weldability of the new family of micro-alloyed steels which have very low carbon contents.

The maximum HAZ hardness has often been proposed as an easily determined indicator of cracking susceptibility. In general, the susceptible microstructures are those with a maximum hardness greater than 300 VHN.(26) However, with a fuller understanding of the factors that contribute to delayed cracking, it has become apparent that HAZ hardness alone does not provide an index of crack sensivitity. Thus cold cracking susceptibility should be related directly to the microstucture itself. Detailed investigations on the effect of microstructures have shown that twinned martensite is more sensitive to cracking than slipped martensite, bainite composed of fine ferrite laths have intermediate susceptibility, and slipped martensite and low carbon auto tempered martensites are the least embrittled.(29) Increasing prior austenite grain size was also found to increase the embrittlement.

C. Effect of Stress

The stresses in a weldment, arise, basically, from two sources:

1. Thermal stresses due to internal restraint,
2. Reaction stresses due to external restraint.

The differential expansion and contraction of the joint because of the extreme thermal gradients inherent in welding give rise to local strains in weld metal and HAZ which are inhibited by the surrounding continuum, leading to residual stresses in the weldment.(30) The magnitude of the peak local stress and strain can approach the ultimate strength of the materials, although the average can be in the plastic or even elastic range, depending upon the yield strength of the material. The final magnitude of the stress is a function of the length of the weld, welding sequence, electrode and joint geometry. The thermal stresses tend to be maximum in those regions which experience the highest temperatures and are the last to cool. Hence, high longitudinal tensile stresses are produced at regions in or near the weld and compressive stresses at regions away from the weld.(30)

For welded joints under external restraint, the reaction force induced, called the "intensity of restraint", enables weld crack initiation to be prediced without a detailed knowledge of the stress and strain history at the point where cracking is expected.(31) The intensity of restraint is defined as "the magnitude of force per unit weld length necessary to give a unit displacement across the weld." Thus the intensity of restraint is a measure of the degree of stiffness exhibited by the structure during welding. The most significant characteristic of hydrogen assisted cracking is the existance of a critical stress (applied or residual) below which cracking will not occur. This has been evident in all weldability tests developed to study delayed cracking. At stresses well above the critical stress, cracking can occur almost immediately, but at lower stresses, there is an incubation time associated with the cracking, giving rise to the term delayed cracking.

The stress that acts as a driving force for hydrogen cracking is, in most instances the welding residual stress, although in exceptional cases, transients such as thermal stress may play a part. In conditions of complete restraint, the residual stress is equal to the yield strength, but in most instances, the structure relaxes elastically to some degree so that the residual stress is lower than yield. In general, longitudinal and transverse residual stresses dominate over through thickness stresses.

D. Mechanism of Hydrogen Assisted Cracking

Several theories have been proposed to explain the mechanism of hydrogen assisted cracking. While the conditions required for hydrogen induced cracking are established, as seen from the above discussion, the exact mechanism of cracking is still not so well understood.

The various theories that have been proposed are the following:

1. Planar pressure theory
2. Surface adsorption theory
3. Lattice interaction theory
4. Lattice decohesion theory

5. Mechanism based on modification of dislocation mobility.

The planar pressure theory of Zapffe and Sims(33) states that molecular hydrogen can precipitate into internal voids in the lattice, thereby building up pressure sufficient to cause premature fracture. Due to straining, the voids enlarge, resulting in a decrease in hydrogen pressure. For further damage to occur, continued preciptiation of hydrogen at the voids is necessary. If the rate of straining is greater than the restoration of necessary pressure, embrittlement should decrease. The effect of lowering the test temperature should be the same, because in this case, the rate of hydrogen delivery into the void and gas pressure inside the void would both decrease.

The surface adsorption theory of Petch and Stables(34) suggests that the hydrogen already in a metal lattice diffuses to the crack tip and reduces the fracture stress by adsorbing on the crack surface. This model was based on Griffith's criterion using energy considerations and thus plasticity effects were ignored.

Troiano proposed the "lattice interaction theory" in which the "true fracture strength" of the lattice was lowered by hydrogen diffusing under the influence of a stress gradient to regions of high triaxial stress within the lattice, causing a loss in lattice cohesion.(35) Oriani and Josephic proposed a theory called "lattice decohesion theory" based on Troiano's model.(36) The decohesion theory suggests that the maximum cohesive strength is reduced by an amount proportional to an undefined function of the hydrogen content at the crack tip. The crack grows when the local tensile elastic stress normal to the plane of the crack equals the local cohesive force per unit area as reduced by the large concentration of hydrogen there. The velocity of the crack growth is controlled by the kinetics of hydrogen transport to the highly elastically stressed region.

Beachem was the first to propose a new model for hydrogen assisted cracking that deviated from the hitherto held concept that hydrogen enhances separation of atomic bonds by a non-deformation type decohesion mechanism.(37) His theory does agree with the earlier proposals that dissolved hydrogen lowers the true fracture strength of the lattice, if this is extended to mean that dislocation motion and multiplication are made easier. The model suggests that instead of hydrogen "locking" dislocations in place, it "unlocks" them and allows them to multiply or move at reduced stresses. Based on the presence of microscopic plasticity in the fracture morphology, Beachem proposed that the presence of sufficiently concentrated hydrogen dissolved in the lattice just ahead of the crack tip aids in whatever deformation processes the microstructure will allow. Intergranular, quasicleavage or microvoid coalescence fracture modes will operate depending upon the microstructure, the crack tip stress intensity, and hydrogen concentration.

Considerable controversy exists today on the validity of one or the other theories presented above.(15) There has been evidence presented to prove or disprove the various theories. While there appears to be no single theory that can explain all the experimental results obtained, Beachem's model of hydrogen-dislocation interactions seems to be the most acceptable. This has been confirmed by a few recent investigations on the occurrence of hydrogen induced cracking in steel weldments.(38,39)

E. Sources of Hydrogen in Flux Cored Wires

As seen from the discussion so far, presence of a sufficient concentration of hydrogen is a necessary condition for hydrogen induced cracking. Consequently, it is necessary to know if flux cored arc welding does engender sufficient hydrogen in the weld atmosphere and if it does, what the possible sources of hydrogen are.

Data published by the Welding Institute, United Kingdom, on diffusible hydrogen contents of CO_2-shielded flux cored wires places FCAW in the category of medium to high hydrogen processes with diffusible hydrogen contents in the range of about 7-33 ml/100 gms of deposited metal.(14) Work done by Evans and Baach on carbon dioxide-shielded seamless flux-cored wires indicates that the values can be as low as 1 ml/100 gms of deposit for a basic type wire, provided the wires had been annealed at high temperatures around 700°C during manufacture. They also found that it was more beneficial to heat treat the wires above 500°C than to bake the flux prior to charging in the core. A recent report on Innershield (self-shielded) welding indicates that the diffusible hydrogen content of this wire is of the order of 4-5 ml/100 gms deposit.(41) In contrast to the situation for SMA welds, there is, thus only limited information available on the hydrogen content of flux cored arc welds.

There are two principal sources of hydrogen in welds made with flux cored wires: (i) the lubricant or drawing compound on the wire surface and (ii) the flux components wrapped into the cored wire. The flux cored wires are generally of a tubular construction fabricated by forming a "U" shape from steel strip, filling it up with the flux ingredients and drawing it through dies to close the seam. The wire is further drawn through dies to obtain the desired size. Thus the manufacturing process involves wire drawing and tube forming followed by an annealing heat treatment to remove the effects of cold work and make the wire less brittle. It is obvious that lubricants used during the drawing process can get entrapped in the surface layers.

The moisture content of the flux components is the other possible source of hydrogen. To analyze this possibility some understanding of the combination of core ingredients, known as the slag system, is required. The slag system has both a physical and metallurgical function. During welding the slag ingredients influence the operating characteristic of the electrode and interact with the parent metal to develop its composition and mechanical properties. Silicates, oxides, carbonates, various organics, ferroalloys, fluorites, titanates, carbon compounds and metal powders are the various possible flux ingredients.(5)

Flux-cored electrodes generally fall into two categories-- those based on titanium dioxide (rutile) and those based on calcium carbonate plus calcium fluoride. The rutile electrodes have a basicity index, BI, in the 0.3-0.7 range and are called "acid electrodes". Here, the basicity index is defined as follows:

$$BI = \frac{\%CaO + \%MgO + \%MnO + \%FeO + \%Na_2O + \%CaF_2}{\%SiO_2 + \%TiO_2}$$

The major component in addition to TiO_2 are SiO_2 and MnO. The $CaCO_3/CaF_2$ type electrodes tend to have a basicity index in the range 1.5-2.5 and are called "basic" electrodes. Electrodes using acid and basic slag systems are classified by AWS as T-1 and T-5, respectively.(42)

Some ingredients of the slag system may contain water chemically bonded as the water of crystallization. Others can absorb moisture from the atmosphere. Consequently, the process adopted for the manufacture of flux cored electrodes would have an effect on the amount of moisture entrapped by the fluxes and therefore, their contribution to the hydrogen content in the welds. Thus prior baking of the flux ingredients and the temperature at which the finished wires are annealed, subsequently could affect the final diffusible hydrogen content in the welds.

THE EXPERIMENTAL PROGRAM

Materials

A. Filler Wires

A representative series of filler wires was collected for the tests, covering a range of strengths and flux types. The individual coils were letter coded for identification as shown in Table 1 which also indicates their AWS classification. A solid filler wire and E7010 and E7018 covered electrodes were also used in the test program for comparison. In all cases, the shielding gases used for welding were those recommended by the manufacturer.

B. Steels

A variety of steels was chosen for the test program. The grades, chemical compositions, plate thicknesses and mechanical properties are indicated in Table 2. Preliminary cracking tests were conducted on the high strength steel A517F which is sensitive to hydrogen induced cracking. However, the majority of the tests were conducted on A302B, A36 and A588B steels, which are representative of structural and pressure vessel steels with strength levels that match those of the filler wires selected for the program.

Weld Parameters

Generally, one of two weld parameters was used to weld the various test specimens. The lower heat input of approximately 8KJ/cm was chosen to impose a relatively severe cooling rate in the weld zone. The higher heat input of 16KJ/cm was intended to produce cooling rates on the low side of the typical fabrication range of welding conditions. In each case, the current was set within the recommended range for the electrode diameter and type and travel speed was adjusted to obtain the desired heat input. The weld parameters used to obtain the two heat input levels for each of the wires are presented in Table 1. In all cases, the gas flow was at the rate of 50 cubic feet per hour and the electrode stickout was held at 25 mm (1"). The welding was carried out on an automatic welding machine with DCRP.

Testing Methods

The experimental program consisted of two main parts:

1. Determination of the diffusible hydrogen content of the various filler wires selected for the program by hydrogen evolution tests,

2. Assessment of hydrogen induced cracking susceptibility in welds made with these flux-cored wires by implant and restraint specimen tests.

A. Diffusible Hydrogen Measurement

In recent years, several new techniques have been developed to measure the diffusible hydrogen content of welds.(43,44) All these techniques involve making a weld bead under standard conditions with the electrode to be tested and collecting the hydrogen evolving from the weld in a fixed period of time under a suitable medium. The method that has been accepted by the International Institute of Welding is the one developed by the Welding Institute, England.(14,45) This method, however, uses mercury as the collecting medium. Due to the health hazards associated with mercury, this test is not widely accepted in the United States. Hence, in this test program, the Japanese method for measuring the diffusible hydrogen content as described in JIS Z 3113-1975 was used.(46) The Canadian Standard CSA W 48.7 M1977 is identical to the above Japanese standard. Figure 3 is a photograph of the test set up used.

Testing Procedure

In the initial tests, specimens (127 mm x 13 mm x 16 mm) were cut from a 19 mm thick A302B plate. The welding parameters used were those appropriate for the size of the electrode being tested. (The parameters for "high heat input" listed in Table 1.) Apart from this deviation from the JIS Z 3113, in the size of the specimens some other modifications were used in the test procedure. The previously weighed specimens were clamped in a copper welding fixture similar to the one used in the IIW method (Figure 4).

Immediately after welding, the welding fixture containing the clamped welded specimen was immersed in a cold water bath. In this bath, the welded specimen was unclamped and rapidly transferred to an alcohol/liquid nitrogen bath where specimen quenching continued rapidly to about -70°C.

The above procedure was repeated on two other specimens. Prior to testing, the specimens were washed and dried according to the following three-step procedure established in the IIW method:

1. Rinse in pure ethyl alcohol or acetone for 3-5 seconds.
2. Rinse in anhydrous ether for 3-5 seconds.
3. Dry with a blast of dry argon gas for 20-30 seconds.

The specimens were then introduced into the graduated glycerin burettes in a glycerine bath held at a temperature of 45°C. After 48 hours, the volume of hydrogen collected was noted and the specimens removed, cleaned, dried and weighed to the nearest 0.1 gm. The volume of hydrogen evolved was converted to that at 0°C under 1 atmosphere pressure and expressed as the value for 100 grams of deposited metal.

This procedure was used to measure the diffusible hydrogen content of all the test wires, as well as the solid wire and E7010 and E7018 covered electrodes. The results obtained indicated that the diffusible hydrogen contents were higher than normally expected from these electrodes. At the same time one of the wires (wire K) had been tested by two different investigators and the values they reported were lower than those obtained in this investigation. Hence, it was decided to follow the JIS Z 3113 method verbatim.

The Japanese standard calls for degassing of the steel specimens at 650°C for 1 hour to remove any residual hydrogen in the steel. This was carried out in all the subsequent tests. The specimen dimensions were changed to 12 mm x 25 mm x 130 mm and the specimens after welding were quenched in cold water, cleaned, dried and introduced into the glycerine burettes. The storing in alcohol/liquid nitrogen bath and the three-step washing and drying procedures were eliminated. The diffusible hydrogen contents of all the test electrodes were measured by both the test procedures described above.

In order to determine if the flux cored electrodes are capable of absorbing moisture from a warm, humid atmosphere such as might be encountered on a summer day, the wires were placed for a week in a humidity chamber at 35°C (95°F) and 90% humidity. Care was taken to see that no moisture deposited on the wires, that could cause rusting. The diffusible hydrogen content of these wires was determined. A sample of these wires, exposed to humidity were then baked in an oven at 125°C for 3 hours to see if this would make any change in the level of hydrogen.

In addition to the diffusible hydrogen measurements, specimens of unused flux cored electrodes from some of the wires

(A, D, F, J) were analyzed to determine their total hydrogen potential.

A. Delayed Cracking Tests

Both direct and indirect tests have been used in the past to study the susceptibility of steels to hydrogen assisted cracking.(26) The direct tests involve the deposition of actual welds usually in some form of specimen designed to provide reproducible conditions of severe restraint. The CTS test, the Y groove Tekken test and the Lehigh restraint test are three widely used direct tests that are useful for rapid evaluation of cracking susceptibility. In all three cases, the amount of cracking revealed by subsequent metallographic examination is a measure of the cracking susceptibility. These direct tests have the advantage of testing the full range of both fusion zone and heat affected zone microstructures. Since they simulate actual fabrication joints, their results correlate better with fabrication requirements. However, since the restraint level depends upon the geometries of both the specimen and the weld bead, it is difficult to separate effects resulting from the influence of welding procedure on microstructure from its influence on the effective restraint level.

The indirect tests for delayed cracking involve constant load stress-rupture tests using notched tensile specimens charged with hydrogen, either by heat treating in a hydrogen-containing atmosphere or by electrolytic means.(29) Such tests have the advantage of allowing independent control of stress level, microstructure and hydrogen level. However, the specimens usually simulate the microstructure found at only one point in the weld HAZ. Furthermore, there is some doubt whether hydrogen introduced by this means diffuses in the same manner as hydrogen introduced via a molten weld pool.

(i) The Implant Test

Granjon developed a test called the Implant test that combines certain features of both the direct and indirect tests.(47) A small cylindrical test specimen containing a circumferential notch is pressed into a hole drilled in a base metal "specimen plate". A test weld is then deposited on the top surface of the specimen plate to (i) fuse the top end of the test specimen, (ii) introduce a controlled amount of hydrogen and (iii) create a weld heat affected zone that now contains the circumferential notch, as indicated schematically in Figure 5. The specimen is then loaded in tension and the time to failure noted for a series of tests performed at various levels.

The Implant test has the following distinct advantages:

1. The stress imposed is independent of the welding procedure used.

2. The test specimens are small and simple to machine.

3. The specimen plate can be of a composition different

from that of the specimen, since it does not actively participate in the test.

A major difficulty with Granjon's method lies in locating the notch properly so that it falls within the same HAZ microstructure in each member of a test series. To overcome this problem, Sawhill, et al. developed a modified implant test, incorporating a specimen with a helical notch in place of a single circumferential notch.(48) The entire HAZ is traversed by the helical notch and failure initiates and propagates in the most crack-sensitive microstructure.

In the present test program, Sawhill's modified implant test was used to assess the cold cracking susceptibility of welds made on various steels by the various flux cored wires. The steels were also tested with a GMAW wire and with E7010 and E7018 as a means of comparison.

The 6 mm modified implants with their axes oriented in the rolling direction were machined from the various test steels. The specimens were prepared to slide fit into the hole drilled in the base plate. The center of the hole drilled in the base plate was carefully located to make sure that the axis of the loading arms and the implant were coincidental. The dimensions of the implants and the base plates and the details of the helical notch are presented in Figure 6.

The tests were carried out on a modified creep testing machine (Figure 7). The machine provided dead weight loading with a maximum capacity of 5500 kgs load. It was equipped with an automatic timer actuated at the instant of loading and stopped by a microswitch on the loading grill when the specimen failed.

Before welding, both the implant and the base plate were degreased with acetone. The welding parameters used in the entire test program are presented in Table 1.
Within two minutes of welding, which is approximately the time required for the weld to cool to 125°C, the base plate with the implant was loaded in the test machine. The load was maintained until rupture occurred. If failure did not occur within 24 hours, the specimen was judged unbroken. For each electrode-steel combination a series of specimens were tested to obtain a stress-failure time diagram. From this, the critical stress or static fatigue limit of the electrode-steel combination was determined.

(ii) Lehigh Restraint Test

Despite its obvious advantages, the implant test is often criticized because the intepretation of its data requires a more precise estimate of the acting stress and strain in a weld than can be made at the present time. It is believed that the level of acting stress on a welded joint of a given geometry is influenced both by the parent material yield strength and by changes in the transformation characteristics of the HAZ. So, a test, such as the implant test that requires the knowledge of the residual strain and stress has severe limitations. Only in situations where the steels being tested or compared have similar yield strengths, transformation temperature ranges and volume changes, and thus are likely to develop similar residual stresses, can welding tests with externally applied stress in HAZ be applied. For this reason, until accurate predictions can be made of the residual stresses in a real weld, some people believe that a self restraint cracking test should be used for assessment of the risk of hydrogen cracking.(50)

In order to confirm the results of the implant test in assessing the hydrogen induced cracking susceptibility, a series of Lehigh restraint tests were carried out.

The Lehigh restraint test is a widely established butt-weld cracking test for comparing quantitatively the degree of restraint at which cracking occurs in welds.(26) The specimen is designed, as shown in Figure 8, with slots cut in the sides and ends. By changing the length of the slots the degree of plate restraint on the weld is varied from one specimen to another. This restraint is expressed numerically by the width of the specimen between the bottom of the slots measured in inches or twice the dimension "X" in Figure 8. The threshold degree of restraint for cracking is expressed as that width (measured to the nearest half inch), which is just sufficient to cause cracking.

Several combinations of electrodes and steel plates were tested. In each case, three cross sections were cut from the weld, metallographically prepared and examined in an optical microscope to detect the occurrence of cracks.

RESULTS AND DISCUSSION

Hydrogen Evolution Tests

The results of the hydrogen evolution tests to measure the diffusible hydrogen contents of welds deposited by the various wires are presented in Tables 3, 4, 5 and 6. Table 3 represents the results of the preliminary tests carried out to measure diffusible hydrogen content of the different flux cored wires. The steel used in these tests was not degassed nor did the specimens conform to the dimensions recommended by JIS Z 3113. The specimens, after welding, were quenched in cold water and stored in alcohol/liquid nitrogen bath. Prior to being introduced into the glycerine bath they were given a three step drying procedure as described in the previous section.

The results in Table 3 show considerable scatter for a specific wire as in the case of wire C, for example. The values obtained were also considerably higher than those claimed by electrode manufacturers for wires of similar type.

To ensure that the deviations from the standard procedure were not the cause of these supposedly high values of diffusible hydrogen content, a second series of tests were run with specimens of dimensions specified in the Japanese standard. These were degassed at 650°C for 1 hour prior to welding, quenched in cold water immediately after welding, wiped dry and introduced directly into the glycerin burettes. Storing the specimens in alcohol/liquid nitrogen mixture and washing them in alcohol and ether were avoided. It was also ensured for these tests that the glycerin bath had been heated to 45°C and held at this temperature at least twelve hours prior to testing. The results of these tests are presented in Table 4.

On comparing the diffusible hydrogen content for any of the flux cored wires in Tables 3 and 4, it is at once obvious that the values have dropped significantly. This seems to be true in all cases. There also seems to be less scatter in results of any one test run. The question now arose whether the degassing of the specimens or the change in specimen dimensions was responsible for this change. There was also the possibility that there could be variation in hydrogen potential along different sections of the same spool. Further tests were then carried out on the smaller specimens after baking them at 650°C for 1 hour. These results are presented in Table 5. The tests were limited to only four electrodes. From Table 5, it seems that baking the specimens prior to welding has resulted in a significant reduction of the diffusible hydrogen content of the weld. In the case of the three electrodes tested for the effect of baking, the reduction ranges from 0.8 ml/100 gms to as much as 3 ml/100 gms. A possible explanation could be that the baking treatment drove out any residual hydrogen in the steel which could otherwise have contributed to the weld diffusible hydrogen content on fusion. If this were the case, the reduction should have been by a constant amount. It is obvious that further investigations are required to explain these observations. On the other hand, the change in specimen dimension appears to have had a negligible influence on the values obtained for the diffusible hydrogen content. This study also indicates that if different filler metals are to be conpared for their diffusible hydrogen contents, it is necessary to follow the standard procedure closely, since changes in technique seem to be able to introduce considerable variations.

For the purposes of this investigation, the results presented in Table 4 would be accepted as the diffusible hydrogen contents of the different electrodes.

Table 6 is a presentation of the results of hydrogen evolution tests carried out after exposing the wires to a warm, humid atmosphere in a specially constructed chamber. As can be seen, the increase due to humidification is not large compared to the amounts incorporated in manufacture.

As mentioned in the first section, there are two principal ways in which hydrogen can enter a flux cored arc weld--the lubricant on the wire surface and the flux components wrapped into the cored wire. Hydrogen analyses of unused wires showed a very high hydrogen potential, beyond the range that could be measured in the hydrogen analyses. After chemical cleaning of the wires, values as high as 50 ppm were obtained. A great majority of the drawing compound is obviously vaporized before the wire reaches the arc and is flushed out. It is thus possible that the conditions of welding (stickout, current level, cup spacing, etc.) can influence the proportion of hydrogen reaching the weld metal. The quantity of lubricant on the wire surface could also vary within or among coils. Differences observed between test runs on different sections of the same coil could probably be accounted for by this variation.

As far as the capability of the flux components themselves to supply hydrogen, the responses of the acid and basic wires, represented by the T1 and T5 types respectively are different. The basic wires consistently produced lower diffusible hydrogen levels (1-1.5 ml/100 gms) than did the acid types (5-10 ml/100 gms). An overall observation is that the hydrogen level in flux cored welds can exceed that exhibited in welds deposited by properly dried E7018 electrodes.

Delayed Cracking Tests

A. Implant Test Results

The results of the implant tests on various steel-filler metal combinations are presented in Figures 9-14. Wires A and F which conform to AWS E70T-1 and E70T-5, respectively were the wires tested most thoroughly since they are representative of the two main types of flux cored wires, rutile and basic. The results of implant tests were conducted on A302B, a low alloy steel at the lower heat input level of 8KJ/cm with wires A, F and D E70T-G) are presented in Figure 9. For comparison, the stress-time curves obtained for A302B when welded with E7018 and the GMA wire are also included. As can be seen from Figure 9, the A302B steel shows almost no sensitivity to hydrogen induced cracking when welded with E7018 as indicated by the fact that the ratio of the critical stress to tensile stress is nearly 100%. This is in accordance with the fact that the diffusible hydrogen content for the E7018 electrode is very low (0.7 ml/100 gms). On the other hand, the GMA wire, which gave rise to almost the same level fo diffusible hydrogen does seem to cause some sensitivity to delayed cracking, as indicated by a lower critical stress. Wires A, D and F all show higher sensitivity to cracking as indicated by critical stress to yield strength ratios of 0.6 to 0.8 for these wires. The point to be noted here is that wire F with a diffusible hydrogen content of 1.4 ml/100 gms seems to behave in the same manner as wires D and A with approximately 5.5-6 ml/100gms diffusible hydrogen content. The tests on A36 steel at 8KJ/cm with wires A and F showed a similar behavior (Figure 10) but with somewhat higher ratios of critical stress to yield strength.

More detailed tests were carried out on A302B steel at a higher heat input of 16KJ/cm. Wires A, D, F, G, and H were used for the tests, representing a range of types as well as strength levels and diffusible hydrogen contents of the flux cored wires. Figure 11 shows the results of these tests along with those for E7018 for comparison. As in Figure 9 for the lower heat input the E 7018 electrode shows no sensitivity. Wire H, with a diffusible hydrogen content of 1 ml/100 gm, does show a delayed cracking behavior in the sense that with lowering stress, the time for cracking increases, but the critical stress to tensile strength ratio is almost 1. This should be interpreted to mean that only if the residual stress levels in a fabrication weldment, made with this steel-filler wire combination, are of the order of tensile strength would delayed cracking occur. Since the residual stresses are never higher than the yield strength of the steel, this means that hydrogen-induced cracking is unlikely to occur in this case. The wires A, D, F, and G show a stress-time behavior lying within a band, as indicated in Figure 11. The critical stress to yield strength ratio for the wires range from 0.6 to 0.9, with wire F again showing the lowest value. Wire F, as mentioned before, has a diffusible hydrogen content of 1.4 ml/100 gms. It should also be noted that doubling the heat input level seems to have produced relatively little change in the implant test cracking behavior of steel A302B.

Figure 12 shows the results of tests on A588 steel. It is clearly evident that A588 is considerably less sensitive to hydrogen induced cracking than A302B or A36 steels. Even with a cellulose coated E7010 electrode, and at the lower heat input level, the critical stress to yield strength ratio is 0.9. At the higher heat input this ratio is above 1, indicating no sensitivity to hydrogen assisted cracking. Tests were also conducted with wires A and F. As expected, the critical stresses in these cases are higher than those obtained with E7010. These results thus show that A588 steel is not likely to crack due to hydrogen assisted cracking when welded with the flux cored electrodes being tested. The other point to be noted here is that doubling the heat input level shows considerable improvement in the delayed cracking behavior of this steel. This is probably due to the lower hardenability.

The results of the tests carried out on A517F steel are presented in Figure 13. This figure shows the effect of the difference in diffusible hydrogen contents of different flux cored wires on the critical stress. Even though A517F is a steel of relatively high hardenability, welding with wire H (E100T-5) with a very low diffusible hydrogen content of 1 ml/100 gms results in a critical stress to yield strength ratio of 0.85. However, when the tests were conducted with wire A (diffusible hydrogen content ~5 ml/100 gms), the ratio falls considerably to about 0.55. Thus, it is evident that even small change in the diffusible hydrogen content of the fillers being used can have a significant effect on the cold cracking behavior of high hardenability steels.

On a steel of low hardenability, the level of diffusbile hydrogen content does not seem to have any damaging effect. Figure 14 shows the result of tests carried out on a low hardenability pipeline steel, X-65. As can be seen here, both high hydrogen E7010 electrode and flux cored wire C ($\sim$ 3 ml/100 gms) have a negligible effect on the stress-time to rupture curves.

The general conclusion that can be drawn from all these results is that most of the flux cored electrodes show a lower critical stress in implant tests carried out on various steels than that obtained with properly dried E7018 electrodes. Only wire H, with a diffusible hydrogen content of 1 ml/100 gms of deposited metal seems to approach the behavior of E7018 electrode. A question that remains unanswered is why wire F with almost as low diffusible hydrogen content, causes cracking susceptibility comparable with that obtained for the higher hydrogen flux cored wires, in steels A302B and A36. The explanation may lie in the chemistry and tranformation characteristics of the weld metal deposited by this wire. Further investigations are required to determine if similar behavior is found when welding other hardenable steels. In all of the tests, the flux cored wires performed better than the E7010 covered electrodes.

B. Lehigh Restraint Test Results

The results of the Lehigh restraint tests conducted on A302B steel and A517F steel are presented in Tables 7 and 8, respectively. Since these tests were carried out mainly to check the results obtained with implant tests, only a few of the steel-electrode combinations tested in the implant test series were investigated in this series.

The results of the tests conducted on A302B steels clearly illustrate the higher susceptibility to hydrogen induced cracking of welds produced with wires A and F than those deposited with wire H and E7018 electrode. The cracks observed in tests with wires A and F were all in the HAZ. The sensitivity to hydrogen cracking was so high that cracks were sometimes observed on both sides of the groove in the HAZ (Figure 15). Most of the tests were carried out only at the higher heat input level of 16KJ/cm based on the experience from the implant tests, where significant difference in cracking susceptibility was not observed for changes in heat input for this steel. In the case of wire F, tests were carried out at both heat input levels, and the difference in the critical restraint levels obtained was small. The main point to be noted from these results is that they agree well with the implant test results carried out on this steel. Thus, the E7018 electrode shows very little sensitivity to cracking and the wire H

(E100T-5) with a very low diffusible hydrogen content exhibits cracking only in the weld metal. This is to be expected since in this case the weld metal overmatches the base metal. It is interesting to note that wire F with a much lower hydrogen content than wire A (1.4 ml/100 gms vs. 5 ml/100 gms) shows almost the same level of sensitivity to hydrogen induced cracking. This is again in agreement with the implant test results.

A few tests were also carried out on the high strength A517F steel. Wires B, C and D at the 500 MPa (∿ 70 ksi) tensile strength level, all caused HAZ cracks in this steel. The critical restraints obtainėd are more or less in agreement with the diffusible hydrogen contents of the wires. This is to be expected since the higher the diffusible hydrogen content of the weld, the greater the susceptibility to hydrogen induced cracking. Tests with wires H and J which match the base metal in strength, all resulted in weld metal cracking. In fact, even down to very low restraints, extensive weld metal cracks were found for both these wires. The fact that wire H has only 1 ml/100 gms diffusible hydrogen content and wire J has 9 ml/100 gms did not make any difference. It may be concluded that these cracks are probably purely due to the effect of restraint on a higher strength weld metal.

CONCLUSIONS

Based on the results of this investigation, the following conclusions may be made:

1. While the flux cored welding process is generally assumed to be a low hydrogen process, welds produced with flux cored filler wires were found to have diffusible hydrogen contents of 1-13 ml/100 gms of deposited metal, as measured by hydrogen evolution tests.

2. As a group, the rutile type flux cored wires tended to produce weld metal diffusible hydrogen levels of 5-13 ml/100 gms of deposited metal while the basic flux cored electrodes produced 1-1.5 ml/100 gms of deposited metal.

3. In implant tests, hardenable steels such as A302B and A517F exhibited critical stress levels for delayed failure below their yield strengths at low heat inputs with flux cored electrodes producing weld metal with over 3 ml/100 gms diffusible hydrogen. Increasing the heat input did not produce significant changes in the delayed cracking behavior of these steels. Welding with a flux cored wire of low diffusible hydrogen content of 1 ml/100 gms caused no sensitivity to cracking.

4. For a structural steel with low hardenability such as A588B, some susceptibility to cold cracking was observed in low heat input welds made with various flux cored wires in the implant test, but increasing the heat input completely removed

this tendency.

5. Flux cored wires also induced cold cracking in welds made in A36 steels, if the heat input was low, but hardly any susceptibility was found in a low hardenability pipe line steel such as X-65.

6. Lehigh restraint tests confirmed the behavior observed in the implant tests.

7. The results of this investigation suggest that in the application of flux cored wires to structural and pressure vessel steels, preheating may be needed for the higher carbon steels and for low alloy high strength steels, particularly in sections over 1 inch thick.

8. As far as the testing methods are concerned, it may be concluded that the glycerin method for measurement of diffusible hydrogen content of welds is a simple procedure that can be used to compare different filler metals. However, care must be taken in adhering to the standard procedure strictly, since deviations can result in considerable variability. It is also evident that considerable investigation must be carried out to determine factors responsible for variability before this test can be established as a standard procedure.

As far as the cold cracking tests are concerned, the results of this investigation showed that correlation may be possible between the implant test and the well-established Lehigh restraint test. Further research is required to determine the nature of this correlation. This is desirable on account of the simplicity in specimen preparation and execution of implant tests.

TABLE 1

Filler Metals and Welding Parameters

Electrode	AWS Class	Diameter, mm	Low Heat Input	High Heat Input	Shielding Gas
A	E70T-1	1.5	220A-28V-45 cm/sec	280A-28V-30 cm/sec	CO_2
B	E70T-1	2.4	400A-28V-45 cm/sec	400A-28V-45 cm/sec	"
C	E71T-1	1.5	220-28V-45 cm/sec	280A-28V-30 cm/sec	"
D	E70T-G	1.5	"	"	"
F	E70T-5	1.5	"	"	25% Co_2 in A
G	E71T-1	1.5	"	"	"
H	E100T-5	1.5	"	"	"
J	E101T-1	1.5	"	"	"
K	E81T-1 Ni2	1.5	"	280A-26V-25 cm/sec	"
E7010	E7010	4.7	150A-25V-30 cm/sec	-	-
E7018	E7018	4.7	180A-23V-30 cm/sec	-	-
M	ER70S-3	1.5	300A-30V-30 cm/sec	-	2% O_2 in A

TABLE 1

Filler Metals and Welding Parameters

Electrode	AWS Class	Diameter, mm	Low Heat Input	High Heat Input	Shielding Gas
A	E70T-1	1.5	220A-28V-45 cm/sec	280A-28V-30 cm/sec	CO_2
B	E70T-1	2.4	400A-28V-45 cm/sec	400A-28V-45 cm/sec	"
C	E71T-1	1.5	220-28V-45 cm/sec	280A-28V-30 cm/sec	"
D	E70T-G	1.5	"	"	"
F	E70T-5	1.5	"	"	25% Co_2 in A
G	E71T-1	1.5	"	"	"
H	E100T-5	1.5	"	"	"
J	E101T-1	1.5	"	"	"
K	E81T-1 Ni2	1.5	"	280A-26V-25 cm/sec	"
E7010	E7010	4.7	150A-25V-30 cm/sec	-	-
E7018	E7018	4.7	180A-23V-30 cm/sec	-	-
M	ER70S-3	1.5	300A-30V-30 cm/sec	-	2% O_2 in A

TABLE 2

Chemical Compositions and Properties of Steel Plates

	Steel Grades				
% Element	A517F	A302B	A36	A588	X65
C	0.17	0.19	0.22	0.14	0.10
Mn	0.87	1.24	1.08	1.11	1.37
P	0.008	0.008	0.018	0.014	0.004
S	0.018	0.019	0.036	0.024	0.003
Si	0.22	0.21	0.25	0.35	0.33
Ni	0.84	0.08	-	0.17	0.16
Cr	0.57	0.12	-	0.48	0.02
Mo	0.48	0.44	-	0.01	0.15
Cu	0.26	0.10	-	0.34	0.02
Al	-	-	-	0.054	0.09
V	0.04	-	-	0.05	-
OTHER	0.002B	-	-		0.024Nb
Other Properties					
Thickness, mm	19	19	13	22	19
Yield Str., MPa	815	485	280	390	-
Tens. Str., MPa	875	550	510	590	-
% Elong.	30	30	31	27	-
% Red. Area	51	51	59	72	-

TABLE 3

Results of Preliminary Hydrogen Evolution Tests

(Specimen dimension - 12 mm x 16 mm x 130 mm; Specimen unbaked; welded samples stored in liquid nitrogen and alcohol, dried in acetone and ether)

Electrode	Diffusible H_2 content, ml/100 gms			Avg. ml/100 gms
A (E70T-1)	7.6	8.0	7.9	7.8
B (E70T-1)	9.2	8.5	9.0	8.9
C (E71T-1)	4.6	8.5	11.3	5.11
D (E70T-G)	8.5	9.2	-	9.0
F (E70T-5)	1.8	1.9	2.7	2.1
G (E71T-1)	13.0	11.0	16.0	13.0
H (E100T-5)	2.1	1.8	2.7	2.2
J (E101T-1)	13.0	15.0	10.0	13.0

TABLE 4

Results of Further Diffusible Hydrogen Measurements

(Specimen dimension - 12 mm x 25 mm x 130 mm; steel baked at 650°C for 1 hour; specimens water quenched and transferred directly to glycerin bath)

Electrode	Diffusible H_2 content, ml/100 gms			Avg. ml/100 gms
A (E70T-1)	4.2	4.9	5.3	4.8
	4.9	4.6	5.3	4.9
	5.1	6.3	5.4	5.6
B (E70T-1)	8.3	5.0	7.3	6.8
C (E71T-1)	2.9	3.1	2.2	2.7
D (E70T-G)	5.5	5.3	6.6	5.8
F (E70T-5)	1.3	1.7	1.3	1.4
G (E71T-1)	6.6	7.4	7.7	7.2
H (E100T-5)	0.9	1.0	1.1	1.0
J (E101T-1)	10.6	8.5	8.4	9.0
K (E81T1-Ni2)*	11.9	13.2	13.8	13.0
E7018	0.6	0.9	0.5	0.7
M (ER70S-3)	0.9	0.8	0.7	0.8

*Values reported for K by two other investigators:

1.	8.0	6.8	6.2	Avg. 7.0 ml/100 gms
2.	6.9	6.0	-	Avg. 6.5 ml/100 gms

TABLE 5

Effect of Baking Specimens Prior to Welding and Specimen Dimension on Diffusible Hydrogen Content

Electrode	Specimen Dimensions	Baked/ Unbaked	Results ml/100 gms			Avg. ml/100 gms
A (E70T-1)	12 mm x 16 mm x 130 mm	Unbaked	7.6	8.0	7.9	7.8
	12 mm x 16 mm x 130 mm	Baked	4.3	4.7	5.2	4.7
	12 mm x 25 mm x 130 mm	Baked	4.2	4.9	5.3	4.8
D (E71T-1)	12 mm x 16 mm x 130 mm	Unbaked	10.4	9.3	8.3	9.0
	12 mm x 16 mm x 130 mm	Baked	6.4	6.1	7.6	6.7
	12 mm x 25 mm x 130 mm	Baked	5.5	6.6	5.3	5.8
F (E70T-5)	12 mm x 16 mm x 130 mm	Unbaked	1.8	1.9	2.7	2.1
	12 mm x 16 mm x 130 mm	Baked	1.7	0.95	1.3	1.3
	12 mm x 25 mm x 130 mm	Baked	1.7	1.3	1.3	1.4
K (E81T-1Ni2)	12 mm x 16 mm x 130 mm	Baked	14.3	12.7	13.6	13.5
	12 mm x 25 mm x 130 mm	Baked	11.9	13.2	13.8	13.0

TABLE 5

Effect of Baking Specimens Prior to Welding and Specimen Dimension on Diffusible Hydrogen Content

Electrode	Specimen Dimensions	Baked/ Unbaked	Results ml/100 gms			Avg. ml/100 gms
A (E70T-1)	12 mm x 16 mm x 130 mm	Unbaked	7.6	8.0	7.9	7.8
	12 mm x 16 mm x 130 mm	Baked	4.3	4.7	5.2	4.7
	12 mm x 25 mm x 130 mm	Baked	4.2	4.9	5.3	4.8
D (E71T-1)	12 mm x 16 mm x 130 mm	Unbaked	10.4	9.3	8.3	9.0
	12 mm x 16 mm x 130 mm	Baked	6.4	6.1	7.6	6.7
	12 mm x 25 mm x 130 mm	Baked	5.5	6.6	5.3	5.8
F (E70T-5)	12 mm x 16 mm x 130 mm	Unbaked	1.8	1.9	2.7	2.1
	12 mm x 16 mm x 130 mm	Baked	1.7	0.95	1.3	1.3
	12 mm x 25 mm x 130 mm	Baked	1.7	1.3	1.3	1.4
K (E81T-1Ni2)	12 mm x 16 mm x 130 mm	Baked	14.3	12.7	13.6	13.5
	12 mm x 25 mm x 130 mm	Baked	11.9	13.2	13.8	13.0

TABLE 6

Effect of Humidifying Flux-Cored Electrodes
(Exposure, 1 week at 35°C, 90% humidity)

Electrode	Hydrogen Level (Triplicate Tests) ml/100 gms	
	As received	Humidified
D (E70T-G)	5.8	5.8
F (E70T-5)	1.4	2.0
J (E101T-1)	9	11

TABLE 7

Results of Lehigh Restraint Tests on A302B Steel

Electrode	Heat Input KJ/cm	Hydrogen ml/100 gms	Critical Restraint mms (inches)
Wire A (E70T-1)	16	5.0	89 (3.5)
Wire D (E70T-G)	16	5.8	114 (4.5)
Wire F (E70T-5)	8	1.4	89 (3.5)
	16	1.4	100 (4.0)
Wire H (E100T-5)	16	1.0	120* (5.0)
E7018	8	0.7	165 (6.5)

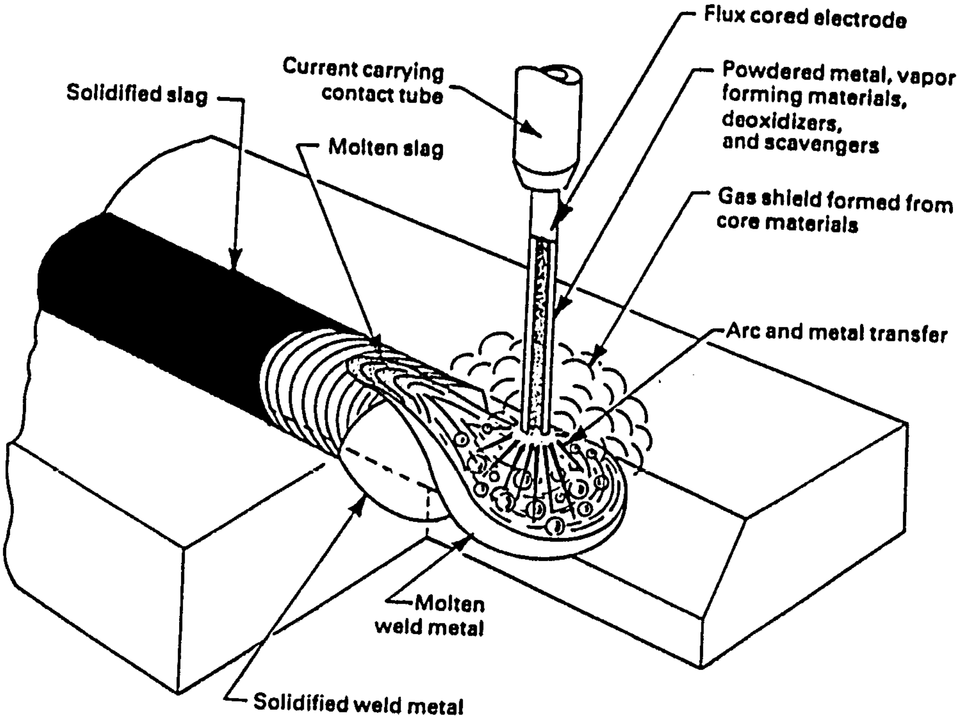
*Cracks in weld

TABLE 8

Results of Lehigh Restraint Tests on A517F Steel

Electrode	Heat Input KJ/cm	Hydrogen ml/100 gms	Critical Restraint mms (inches)
Wire B (E70T-1)	16	7.0	50 (2.1)
Wire C (E71T-1)	16	2.7	90 (3.5)
Wire D (E70T-G0	16	5.8	100 (4.0)
Wire H (E100T-5)	16	1.0	50* (2.0)
Wire J. (E101T-1)	16	9.0	50* (2.0)

*Extensive weld metal cracks obtained even at 2" restraint



Figúre 1. Self-shielded flux cored arc welding.(2)

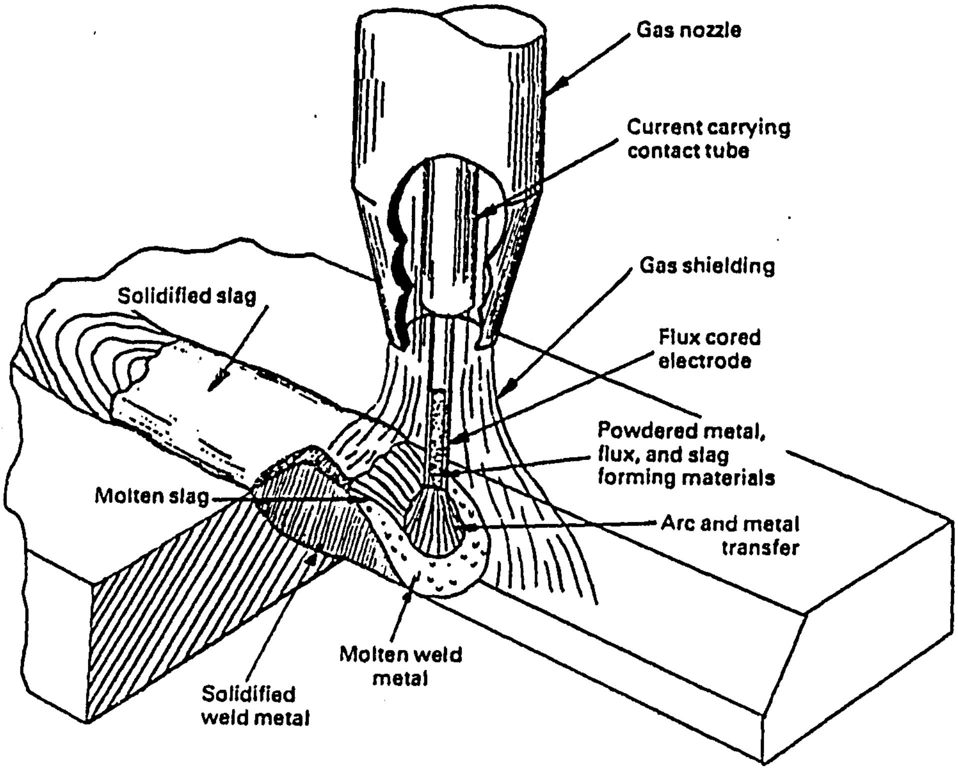


Figure 2. Gas-shielded flux cored arc welding. (2)

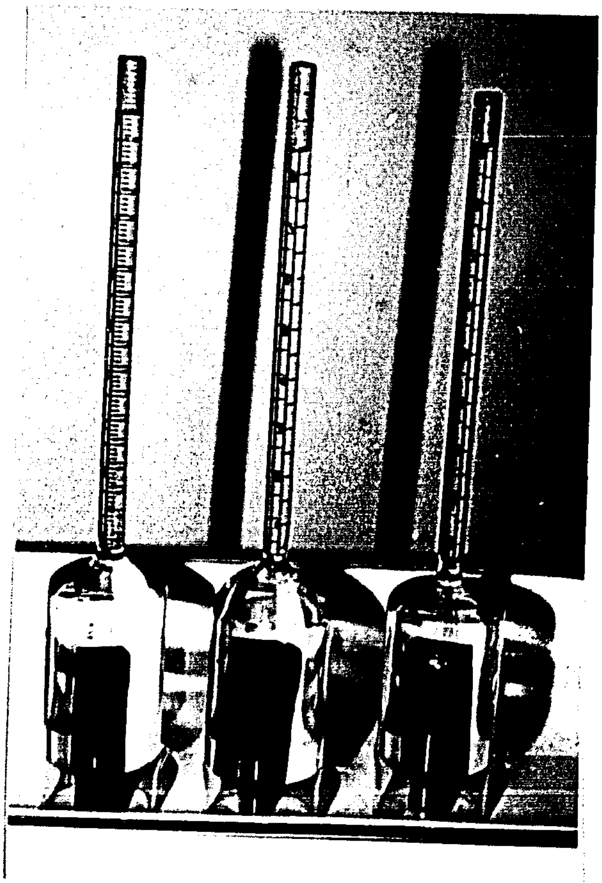


Figure 3. Diffusible hydrogen measurement set-up.

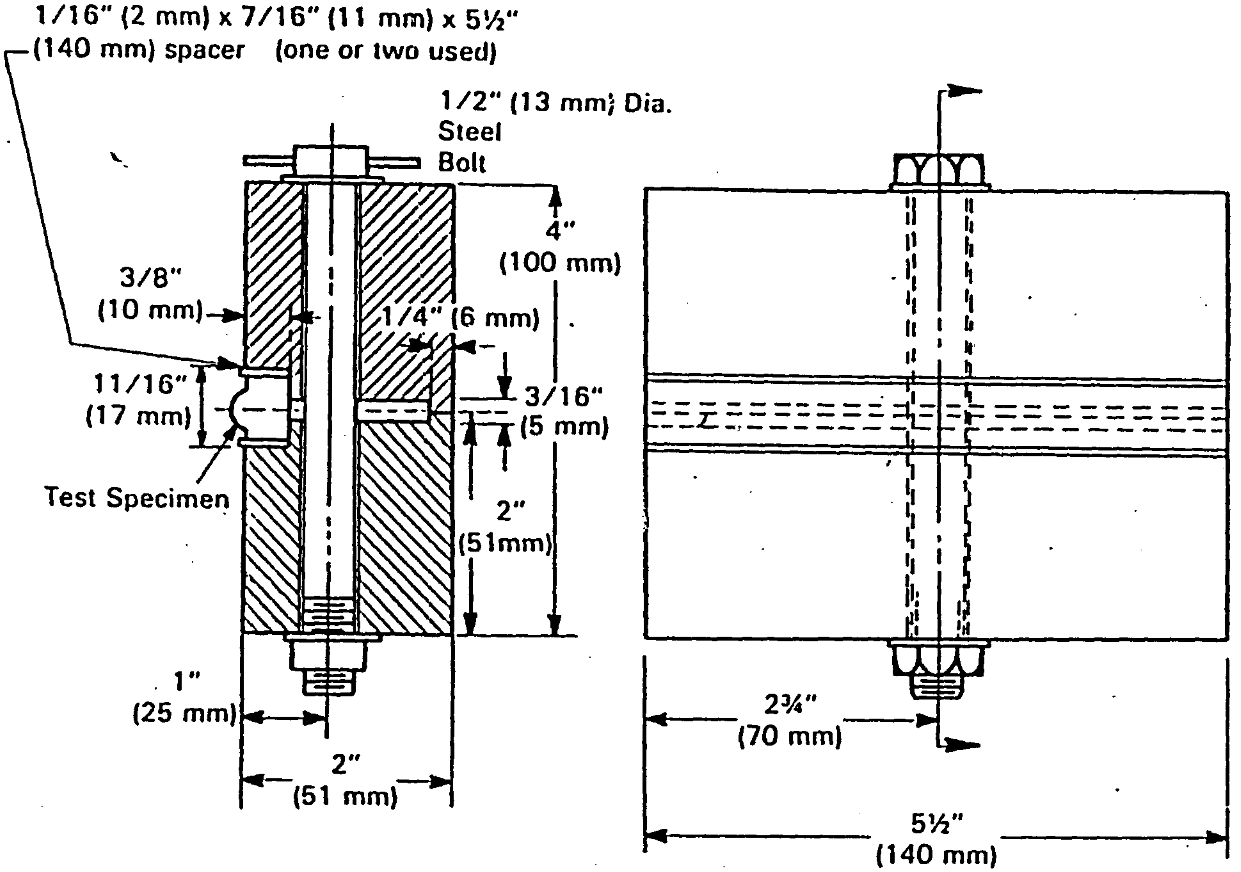


Figure 4. Copper welding fixture.

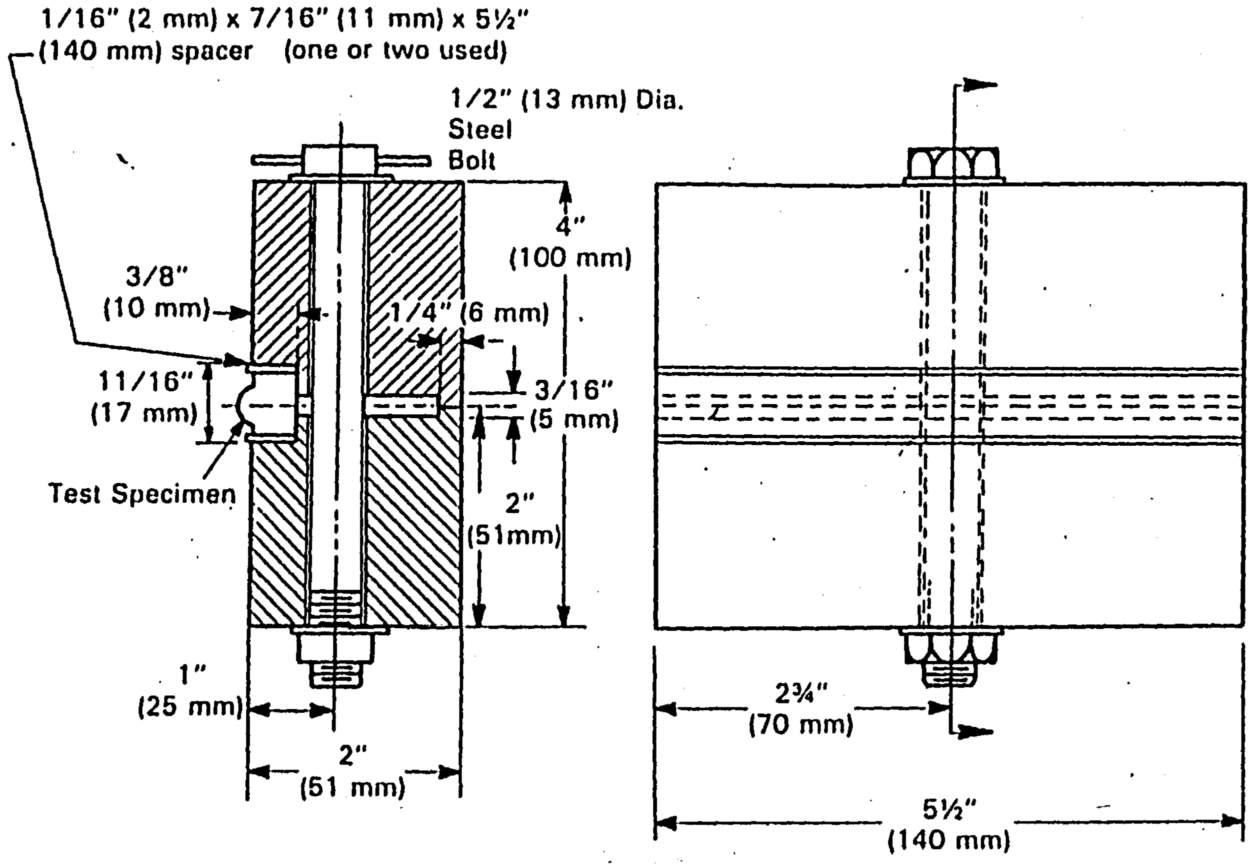


Figure 4. Copper welding fixture.

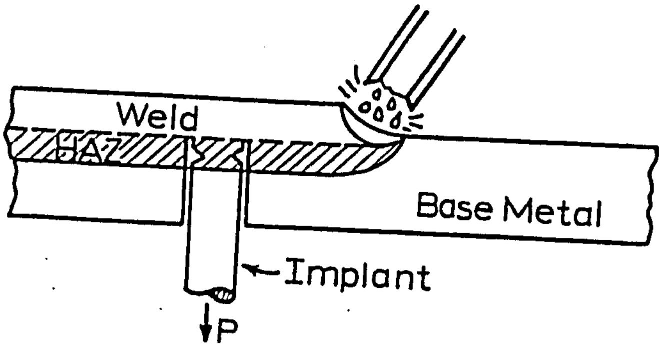
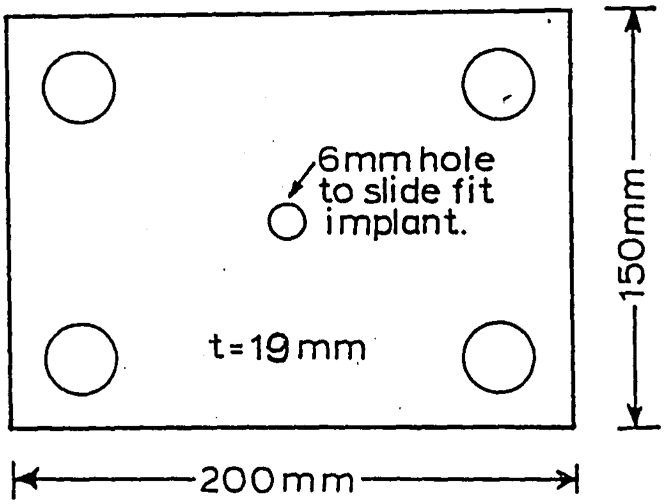
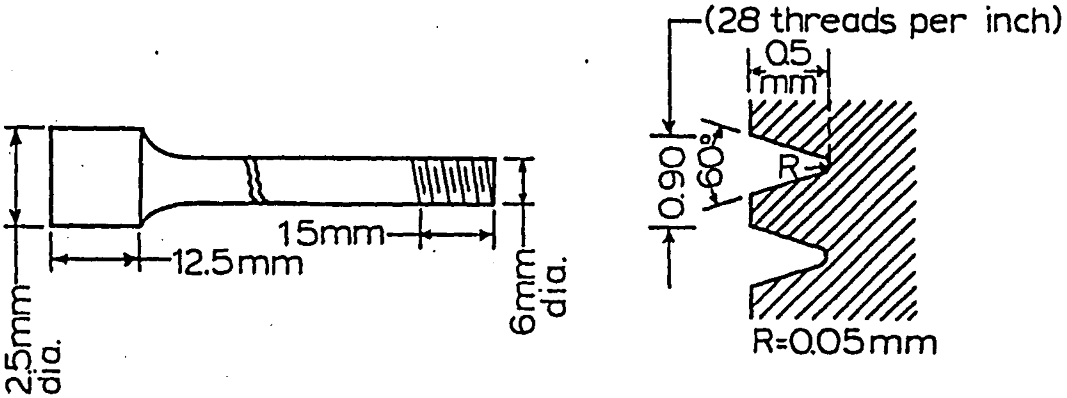


Figure 5. Original implant test.

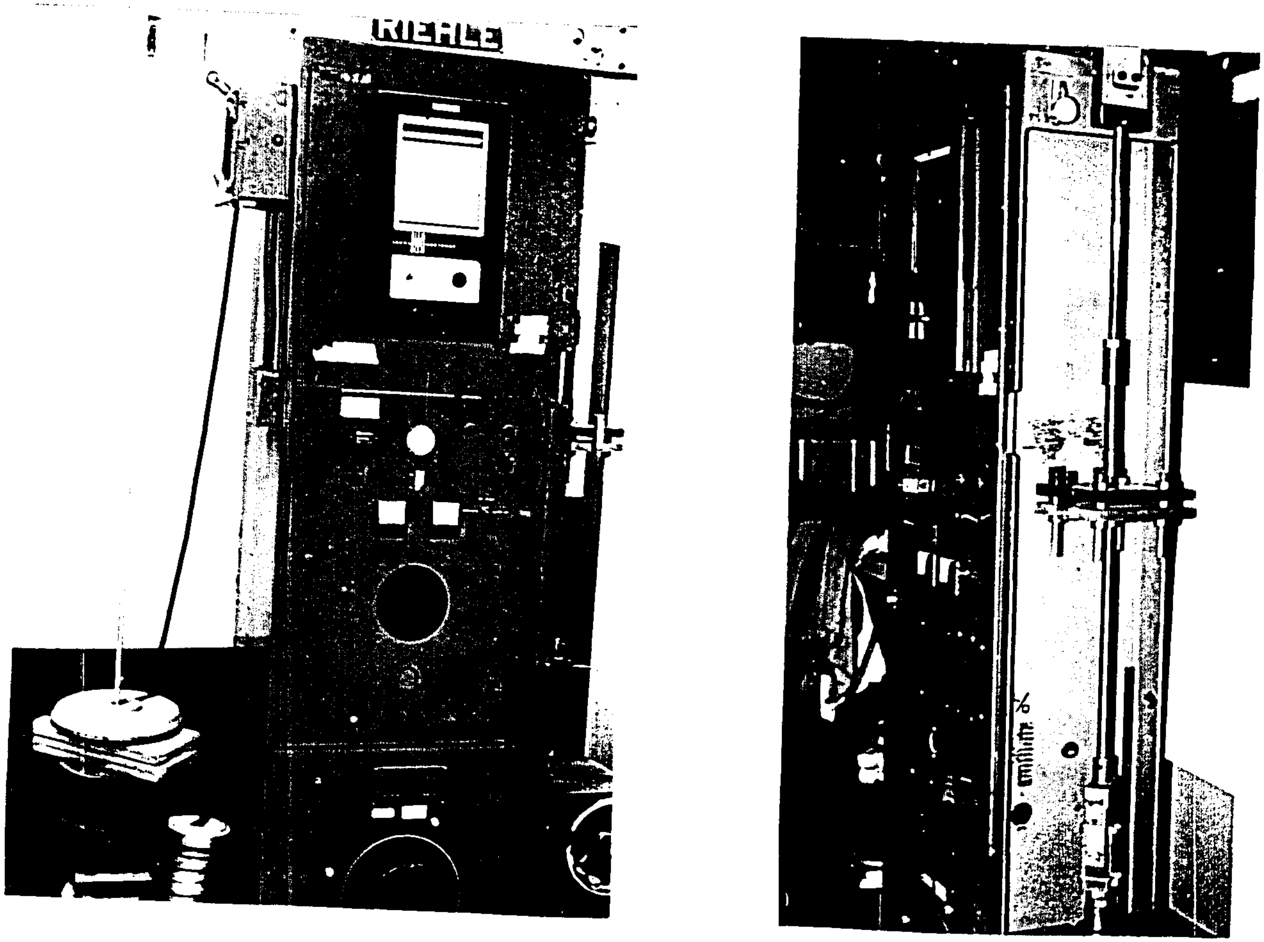


Implant Base Plate.



Implant Test Piece. Details of Helical Notch.

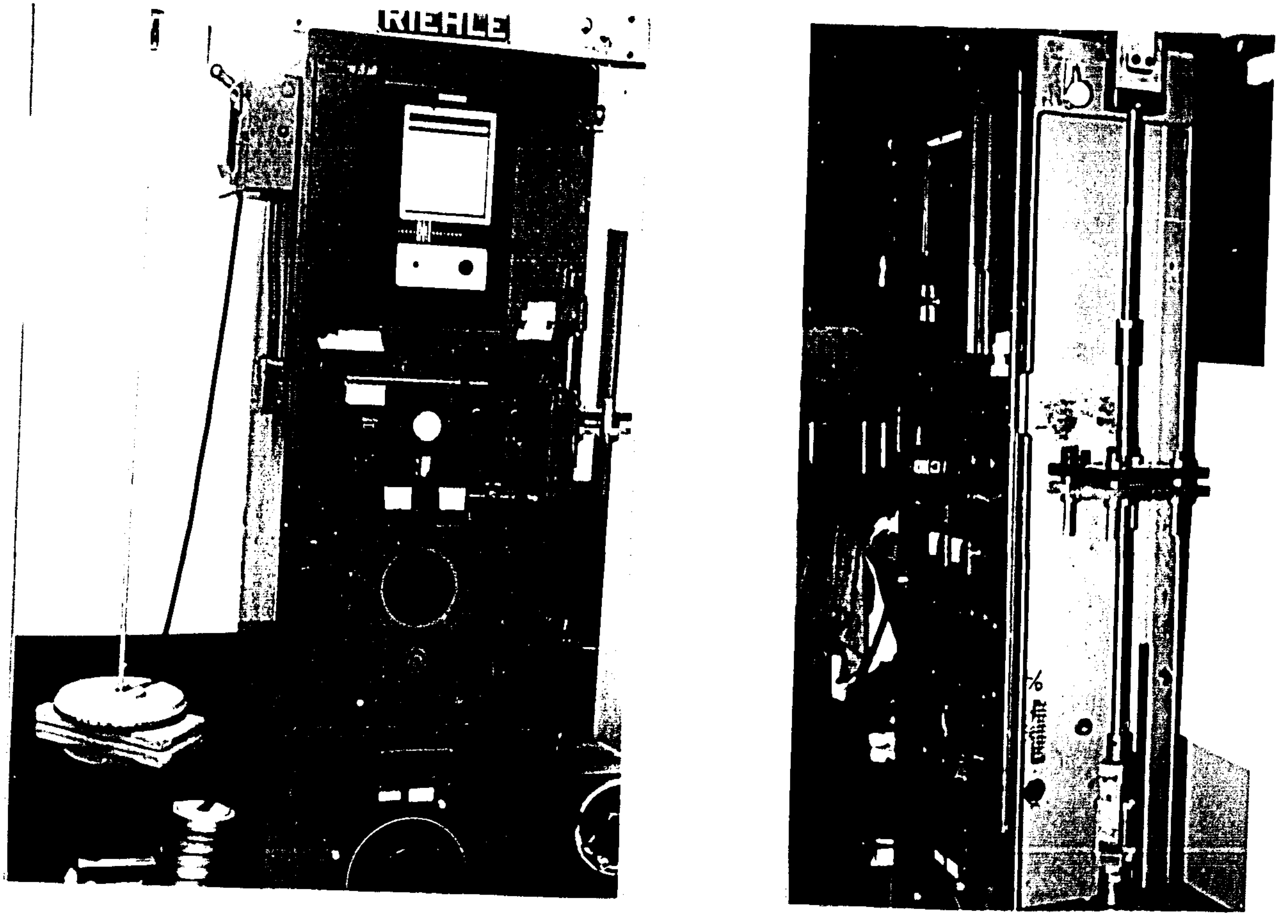
Figure 6. Details of implant specimen and backing plate.



a

b

Figure 7. a) Implant testing machine; b) detail of the specimen loading fixture.



a

b

Figure 7. a) Implant testing machine; b) detail of the specimen loading fixture.

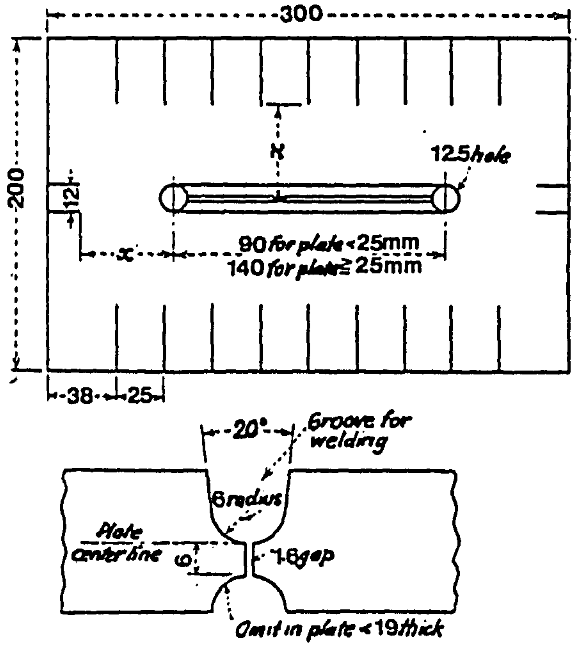


Figure 8. Lehigh restraint test specimen details.

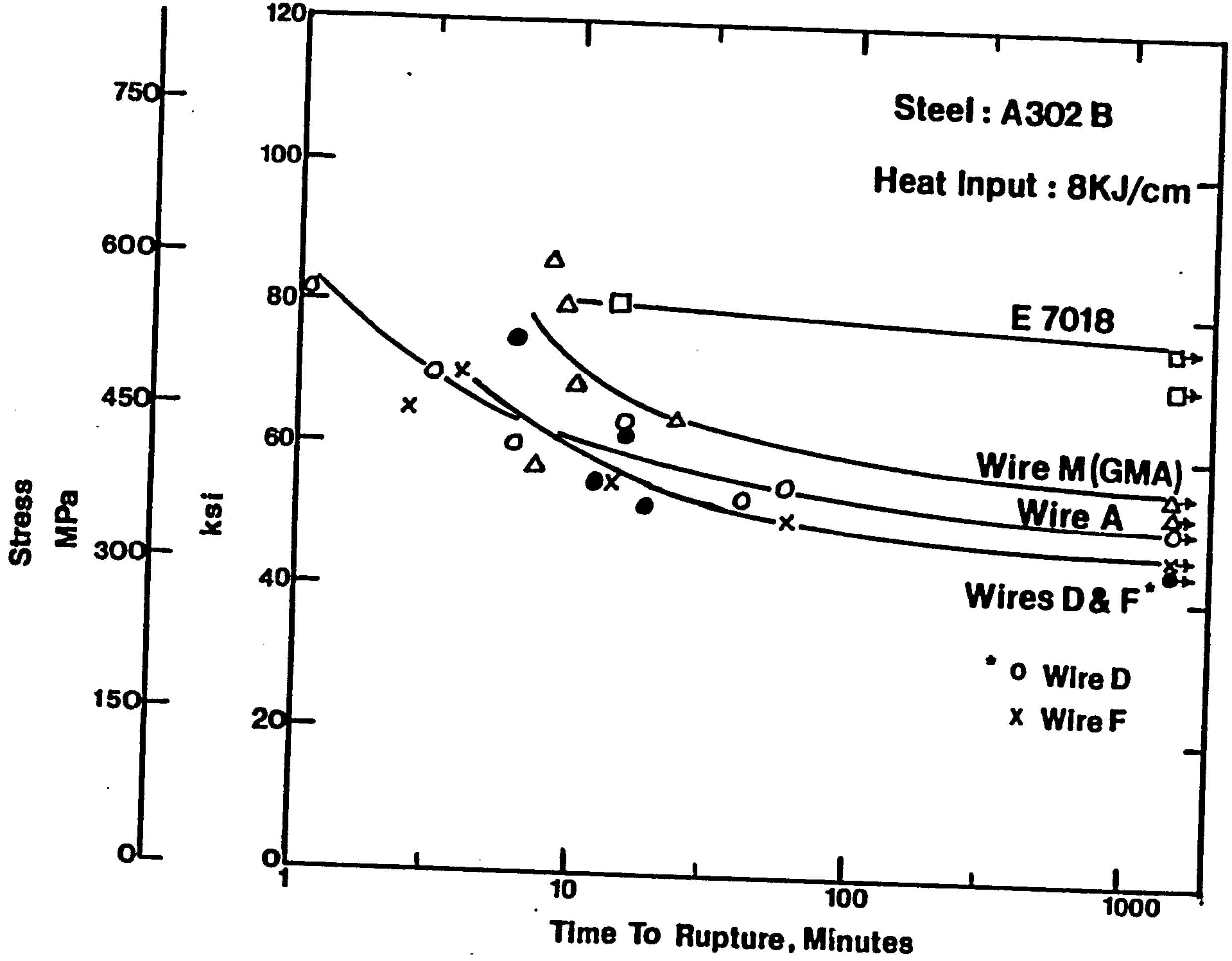


Fig.9 Effect of Filler Wire on Implant Tests of A 302 B Steel welded at 8KJ/cm

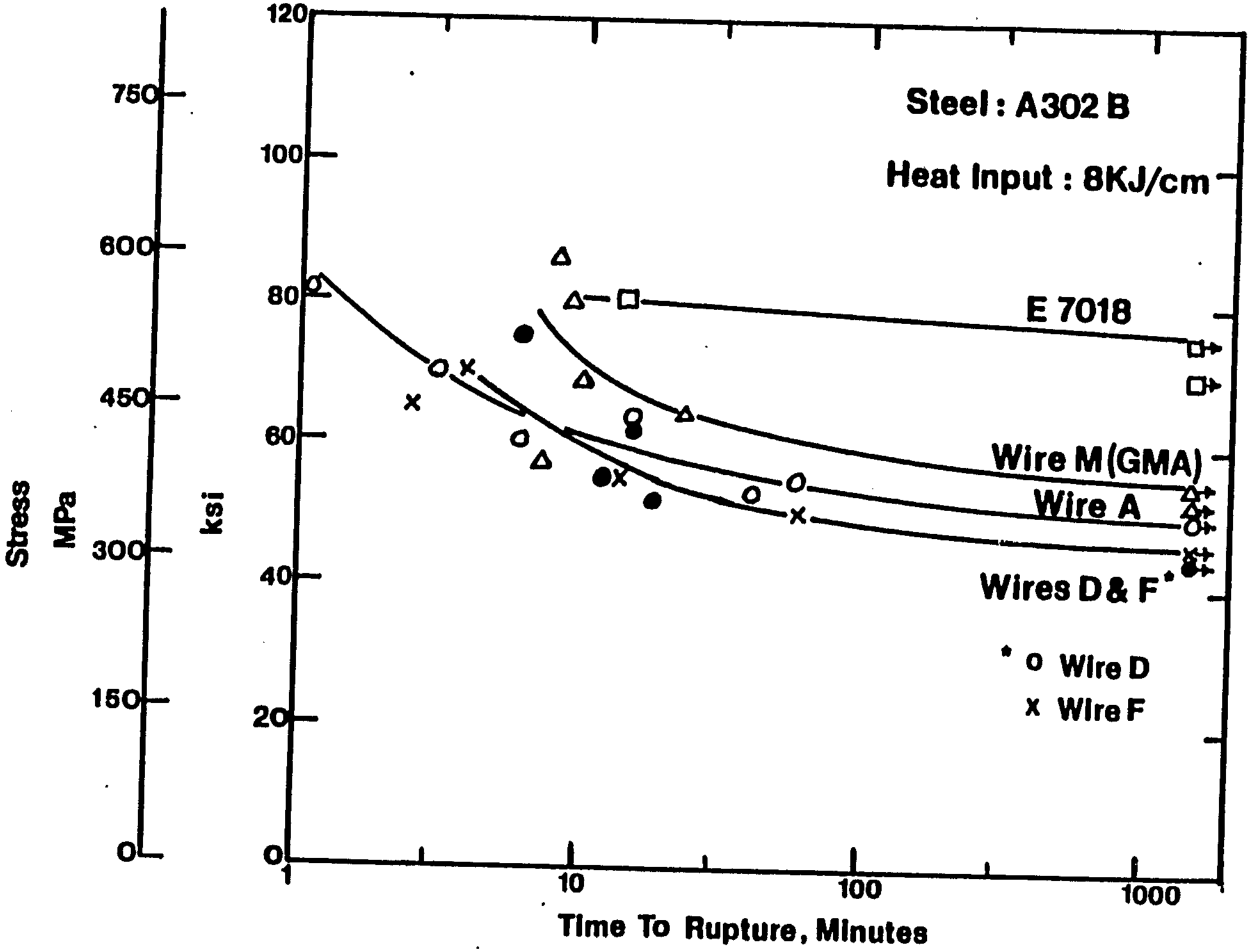


Fig.9 Effect of Filler Wire on Implant Tests of A 302 B Steel welded at 8KJ/cm

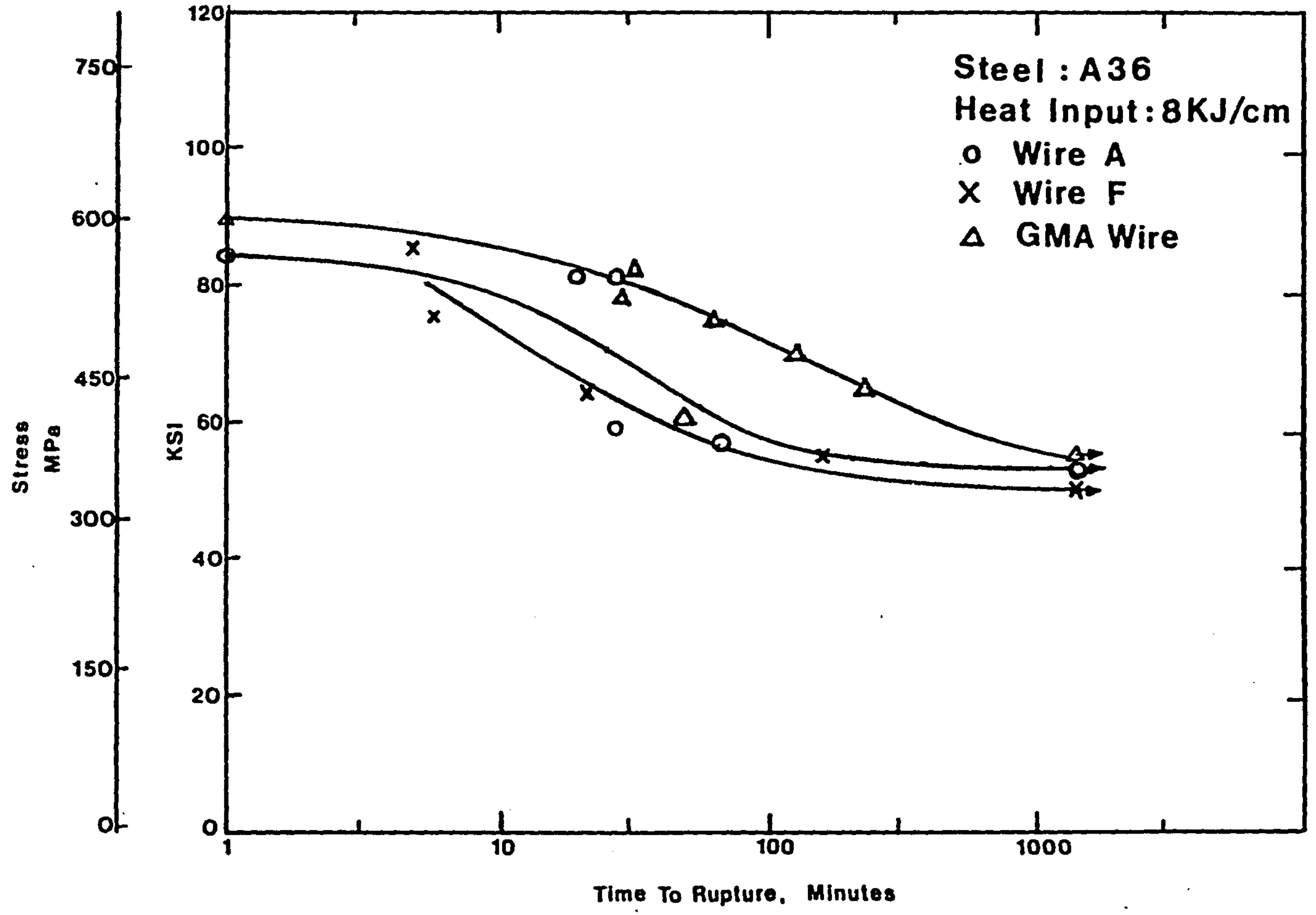


Fig.10 Implant Tests On A36 Steel with Several Filler Wires

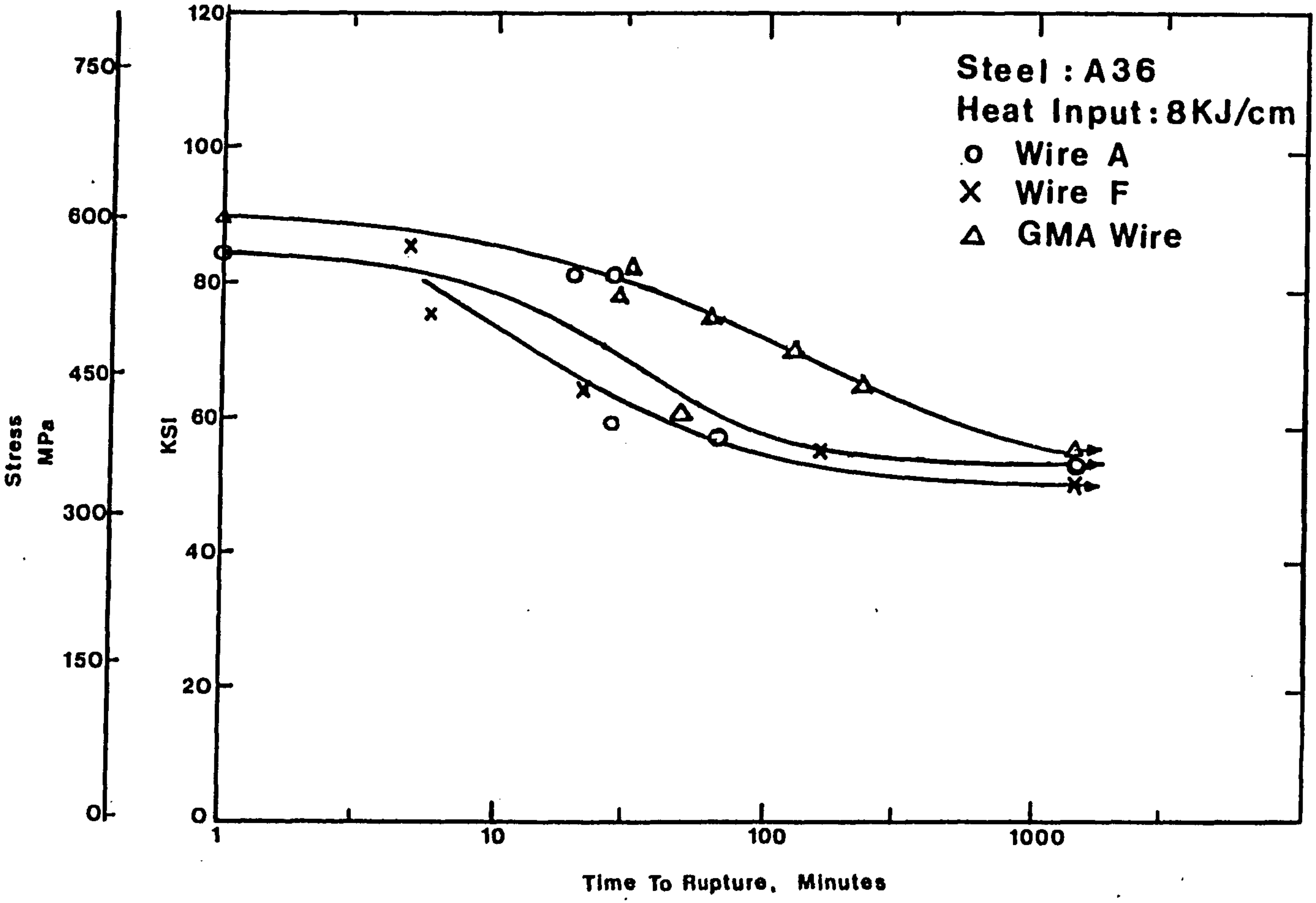


Fig.10 Implant Tests On A36 Steel with Several Filler Wires

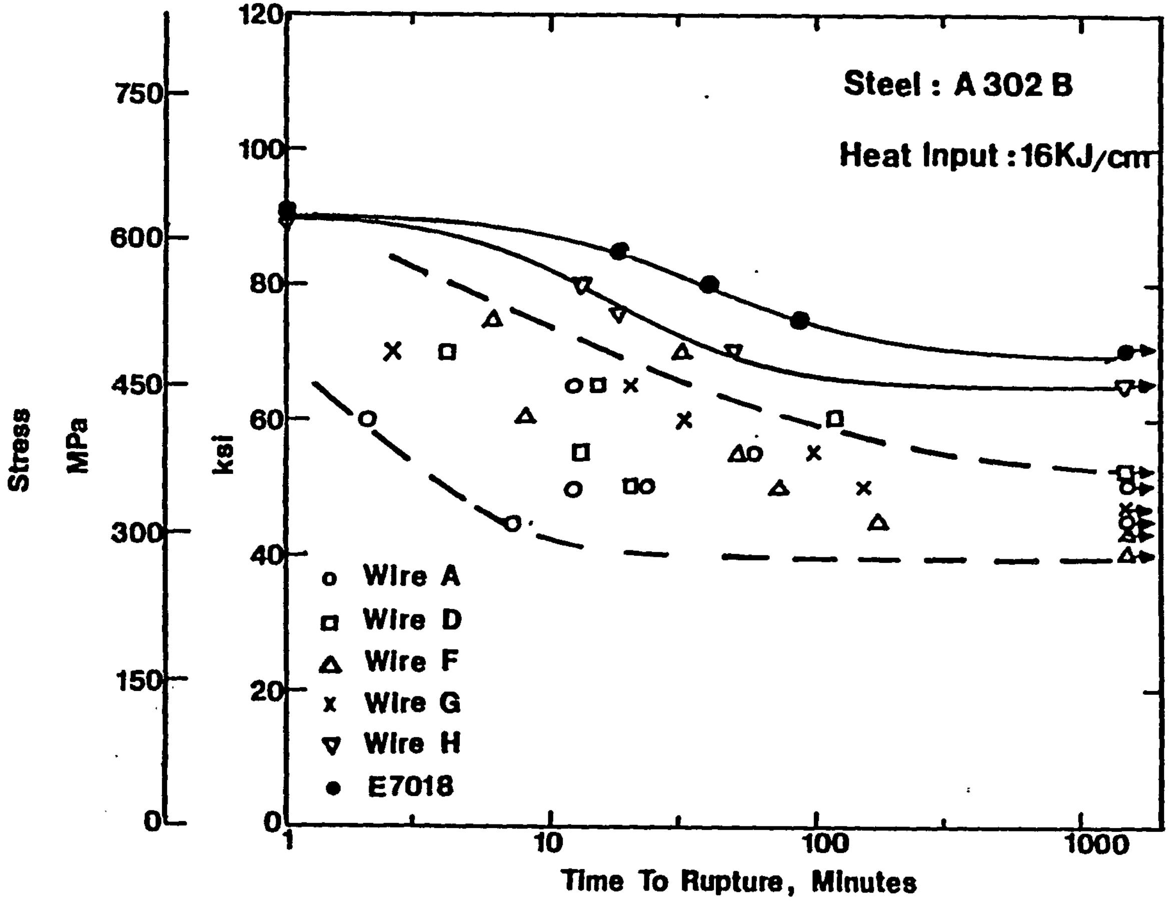


Fig.11 Effect of filler wire on Implant Tests of A 302 Steel welded at 16KJ/cm

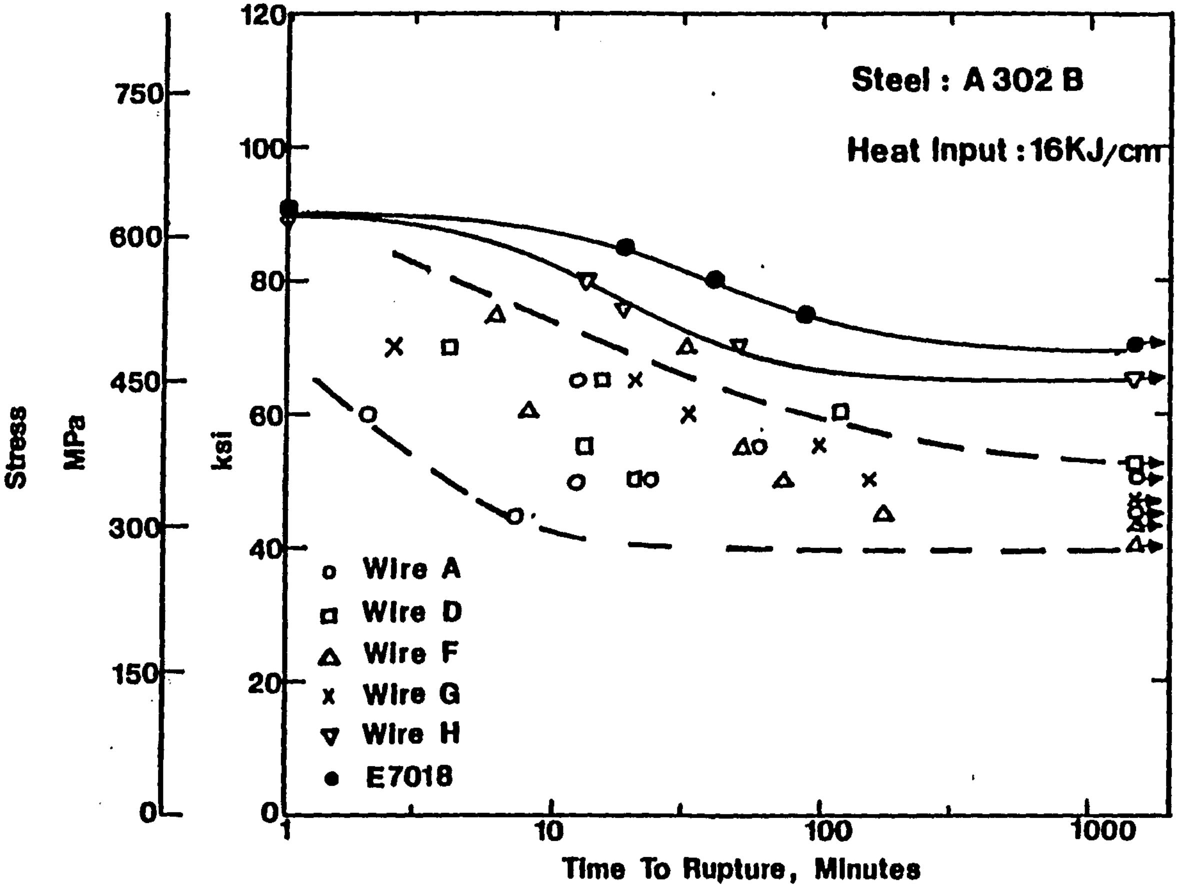


Fig.11 Effect of filler wire on Implant Tests of A 302 Steel welded at 16KJ/cm

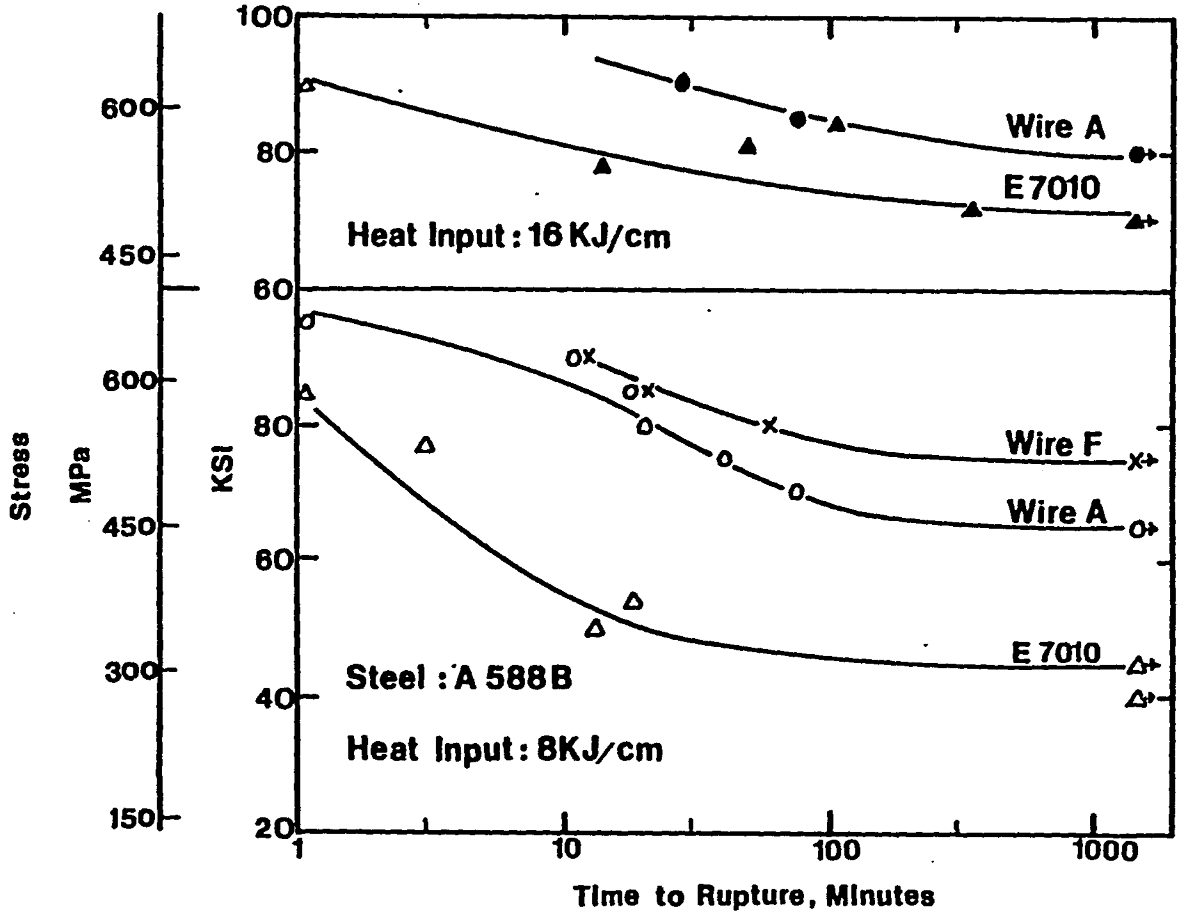


Fig.12 Effect of Heat Input on Implant Test Results for A 588 Steel

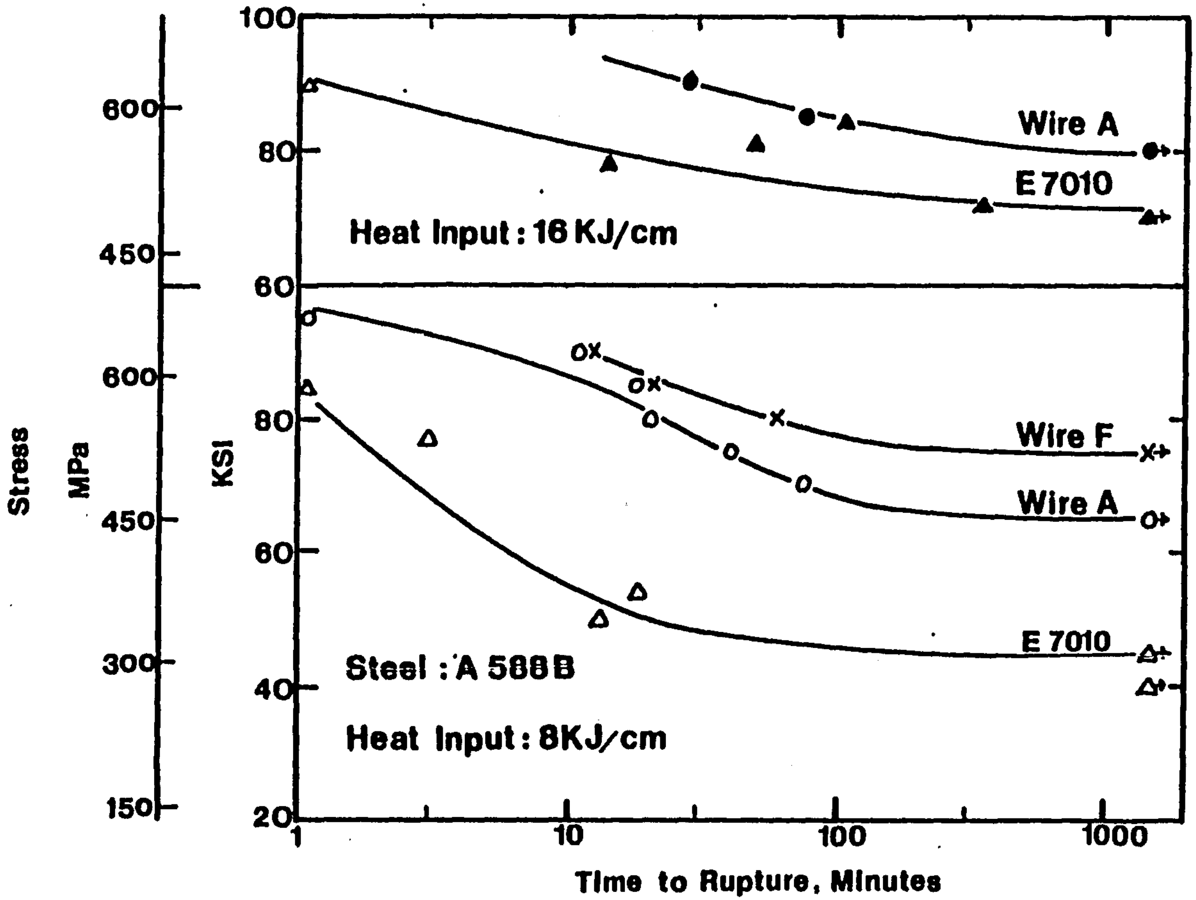


Fig.12 Effect of Heat Input on Implant Test Results for A 588 Steel

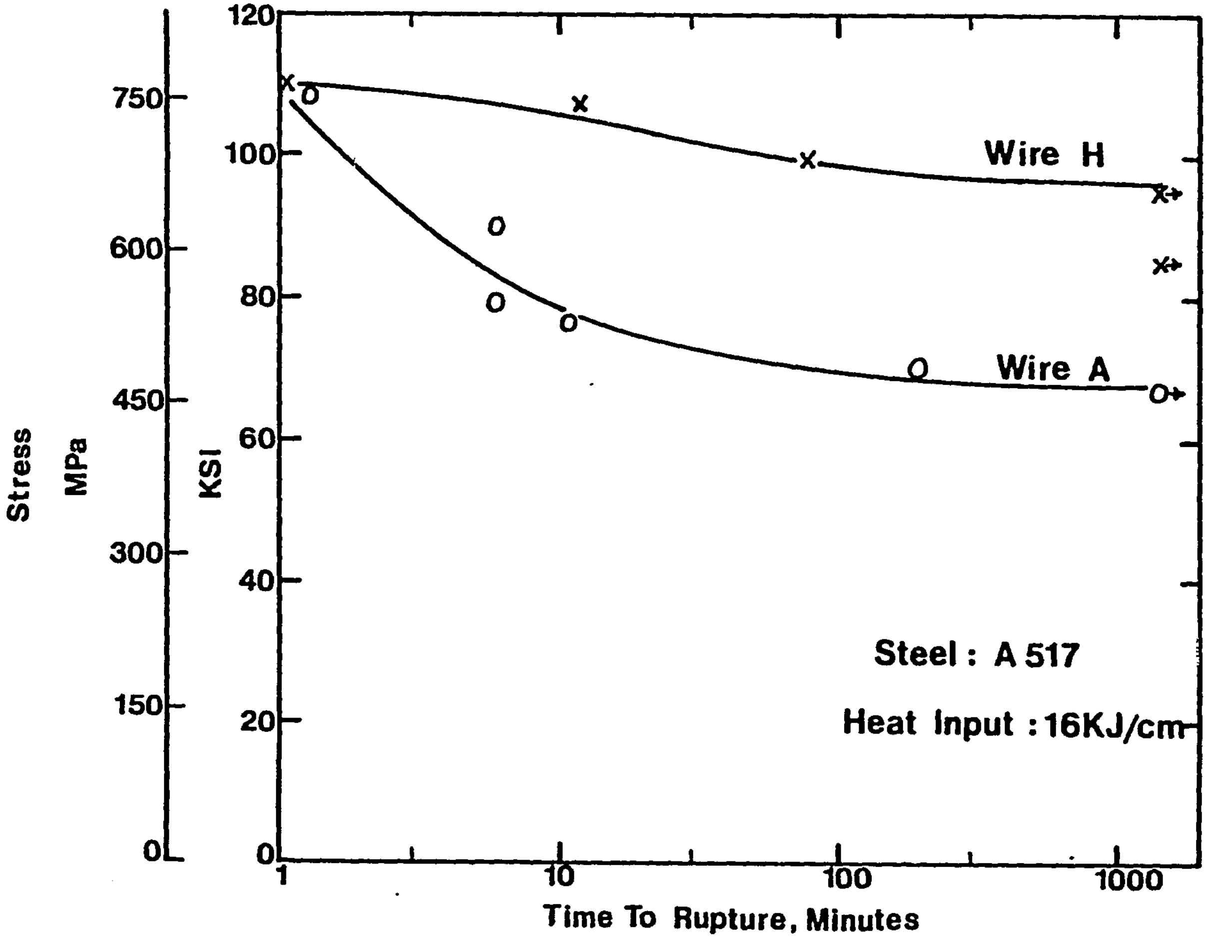


Fig.13 Implant Tests On a Quenched and Tempered High Strength Steel

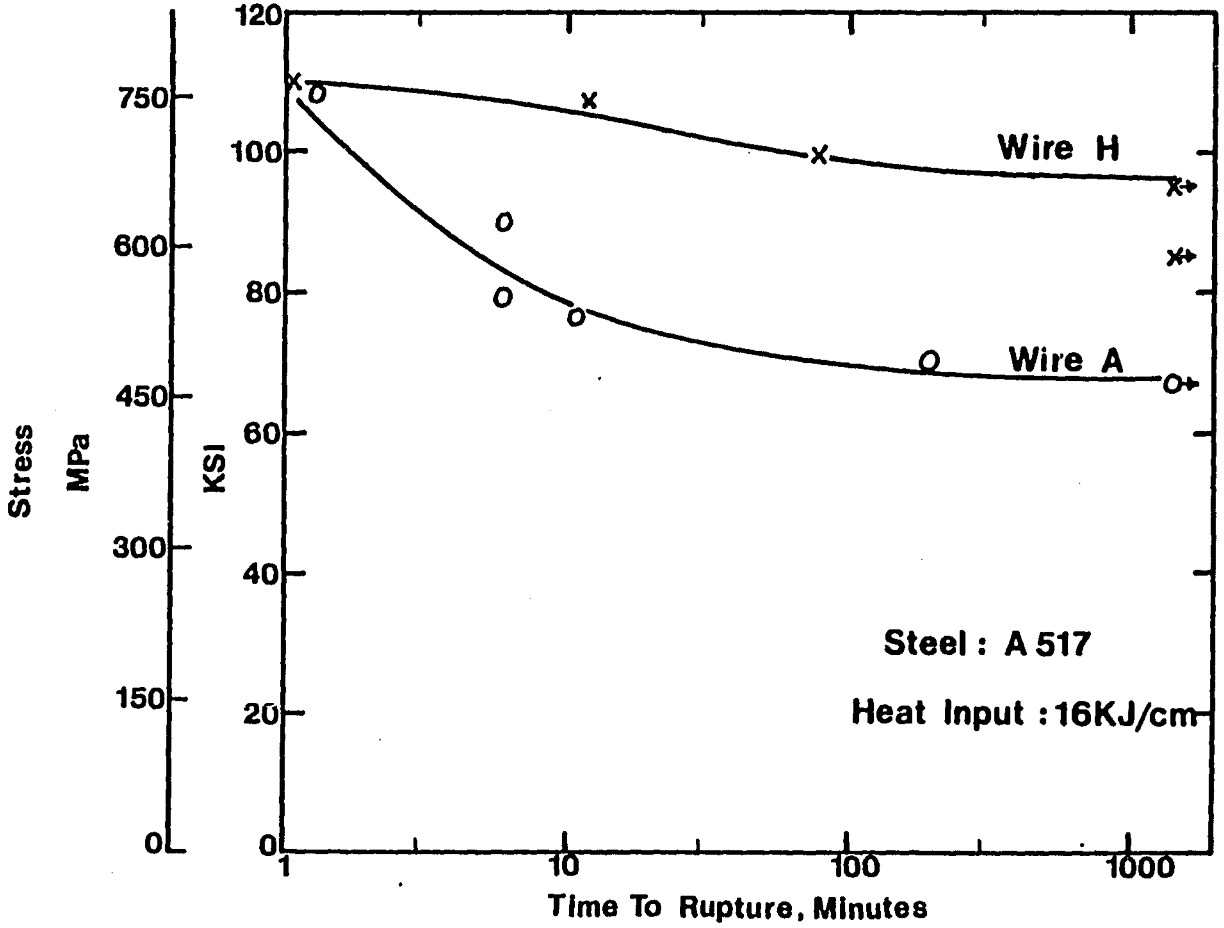


Fig .13 Implant Tests On a Quenched and Tempered High Strength Steel

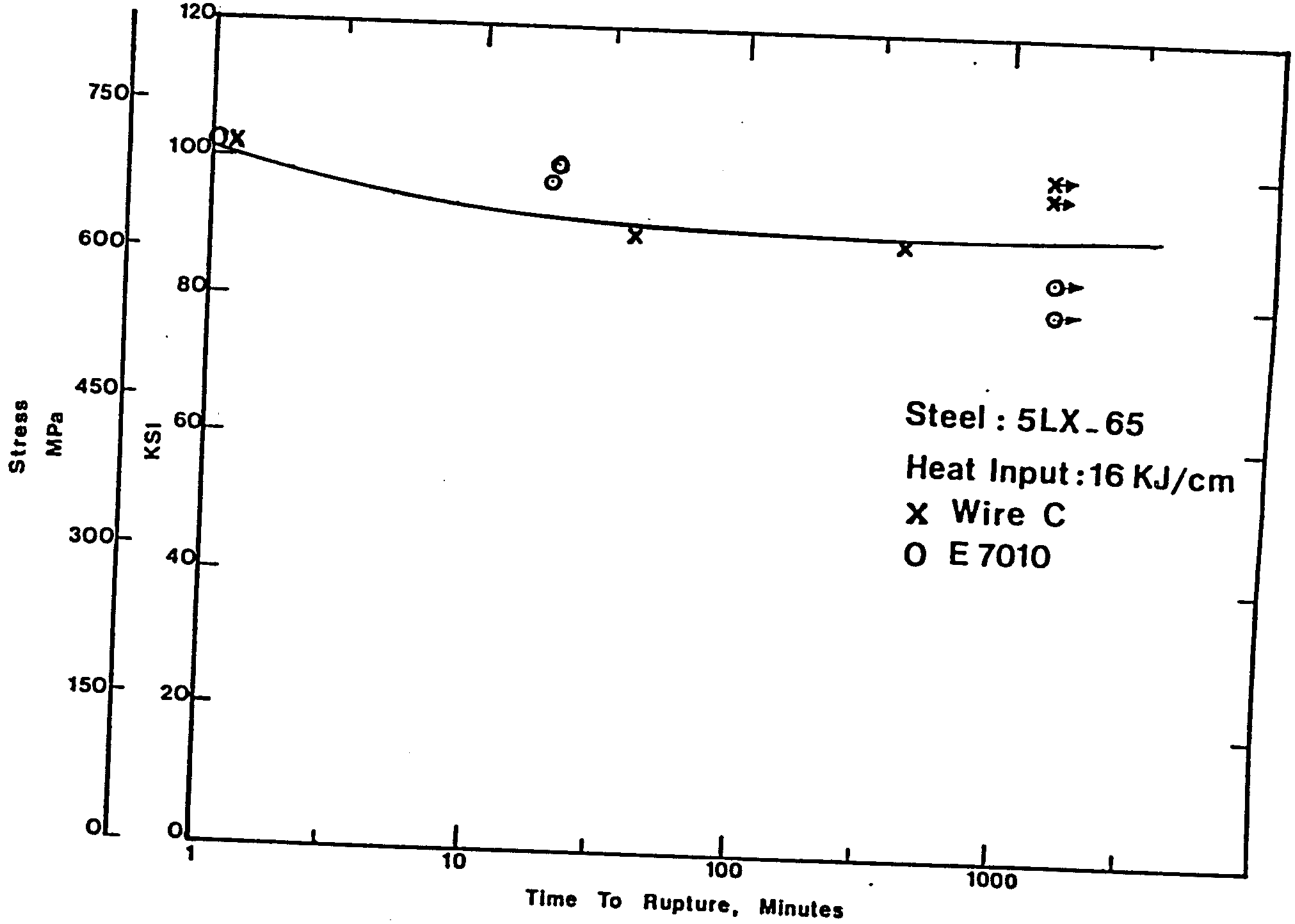


Fig.14 Implant Test Behavior of X65 Pipeline Steel

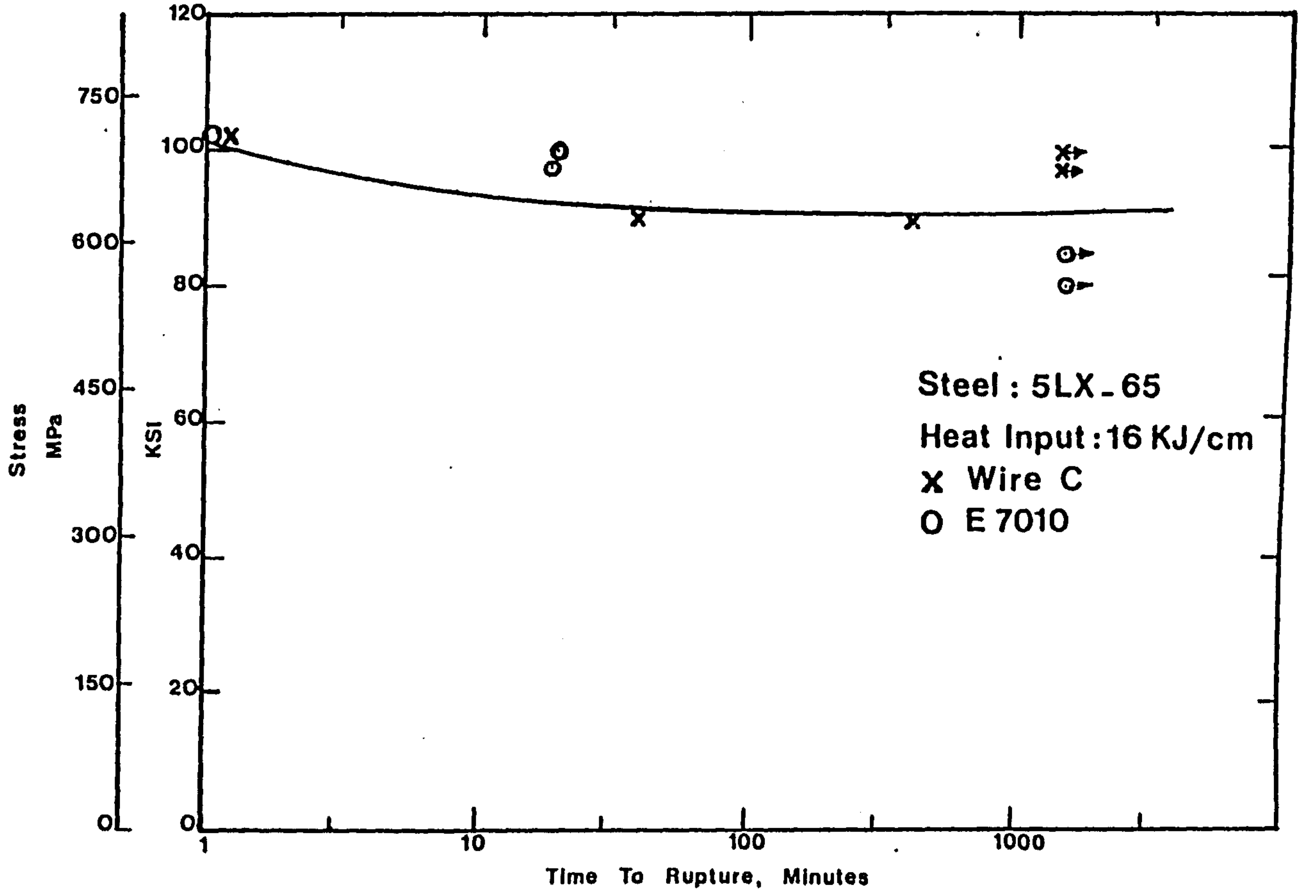


Fig.14 Implant Test Behavior of X65 Pipeline Steel

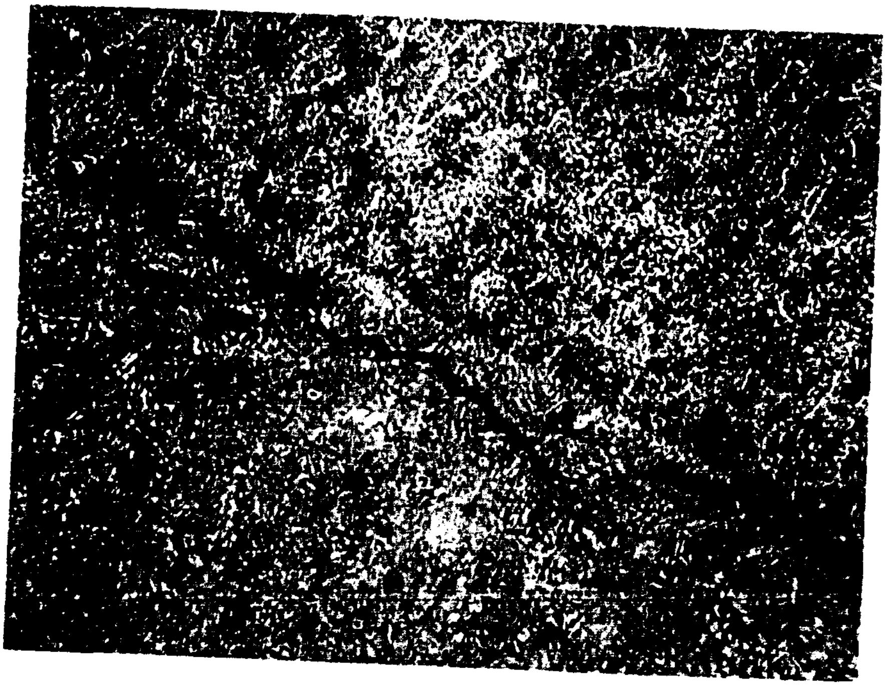


Figure 15. The morphology of HAZ cracks obtained in Lehigh restraint test. 75x

REFERENCES

1. Chouinard, A. F. and Monroe, R. P., "A New CO_2 Welding Process," Welding Journal, Vol. 37, No. 11, November 1957, p. 1069.
2. Welding Handbook, Seventh Edition, Volume 2, published by American Welding Society, Miami, Florida.
3. Weymueller, C. R., "All Position Electrodes Good, Could Be Better," Welding Design and Fabrication, Vol. 55, No. 7, July 1982, p. 49.
4. Weymueller, C. R., "Shipbuilders Modernize to Stay in the Swim," Welding Design and Fabrication, Vol. 55, No. 6, June 1982, p. 68.
5. Siewert, T. A. and Feiree, S. E., "Status Report: Flux Cored Arc Welding," Metal Progress, Vol. 119, No. 5, April 1981, p. 58.
6. Chouinard, A. F. and Howery, J. A., "Progress Report on the Flux Cored CO_2 Welding Process," Vol. 38, No. 4, April 1959, p. 334.
7. Welding Journal, Vol. 52, No. 3, March 1973, p. 155.
8. Hoitomt, M. and Lee, R. K., "All Position Production Welding with Flux Cored Gas-Shielded Electrodes," Welding Journal, Vol. 5, No. 11, November 1972, p. 765.
9. Welding Journal, Vol. 57, No. 5, May 1978, p. 42.
10. Welding Journal, Vol. 57, No. 6, June 1978, p. 36.
11. AWS D1.1 - Structural Welding Code, published by American Welding Society, Miami, Florida.
12. Guide Specifications for Fracture Critical Non-Redundant Steel Bridge Members, September 1978, published by American Association of State Highway and Transportation Officials, Washington, D.C.
13. "Hydrogen in Metals", Proceedings of an International Conference, Champion, PA, ASM, 1973.
14. Coe, F. R., "Welding Steels without Hydrogen Cracking," Welding Institute publication, 1973.

15. "Hydrogen Embrittlement Testing," ASTM STP 543, American Society for Testing and Materials, 1974.

16. "Hydrogen Damage," A metals science source book, American Society for Metals.

17. Sieverts, A, "Die Aufnahme von Gosen Durch Metalle," Zeitschrift for Metallkunde, Vol. 21, 1929, p. 37.

18. Zapffe, C. A. and Sims, C. E., "Defects in Weld Metal and Hydrogen in Steel," Welding Journal, Vol. 19, 1940, p. 377s.

19. Zuzuki, H., Inagaki, M. and Nakamura, H., "Effects of Restraint and Hydrogen on Root Cracking of High Strengths Steel Welds," IIW Co. IX-408-64.

20. Evans, G. M. and Christensen, N., "Correlation of Weld Metal Hydrogen Content with HAZ Embrittlement," Metal Construction and British Welding Journal, Vol. 3, No. 5, May 1971, p. 188.

21. Evans, G. M. and Christensen, N., "Implant Testing--Effect of Steel Composition and Hydrogen Level," Metal Construction and British Welding Journal, Vol. 3, No. 7, July 1971, p. 263.

22. Beachum, E. P., Johnson, H. H. and Stout, R. D., "Hydrogen and Delayed Cracking in Steel Weldments," Welding Journal, Vol. 40, No. 4, April 1961, pp. 155s-159s.

23. Interrante, C. G. and Stout, R. D., "Delayed Cracking in Steel Weldments," Ibid, Vol. 43, No. 4, April 1964, p. 1455.

24. Smith, N and Bagnall, B. I., "The Influence of Sulfur on Heat Affected Zone Cracking of Carbon Manganese Steel Welds," Metal Construction and British Welding Journal, Vol. 1, No. 2, Feb.(s) 1969, pp. 17-23.

25. Moreton, Mrs. J., Coe, F. R. and Bonizewski, T., "Hydrogen Movement in Weld Metals," Ibid, Vol. 3, No. 5, May 1971, pp. 185-187.

26. Stout, R. D. and Doty, W. D., "Weldability of Steels," 3rd Edition, Welding Research Council, New York, 1978.

27. Ito, Y. and Bessyo, K., "Weldability Formula for High Strength Steels," IIW Doc. No. IX-576-68.

28. Winterton, K., "Weldability Prediction from Steel Composition to Avoid Sleat Affected Zone Cracking," Welding Journal, Vol. 40, No. 6, June 1961, p. 253S.

29. Boniszewski, T. and Watkinson, F., "Effect of Weld Microstructures on Hydrogen-Induced Cracking in Transformable Steels," Part I and Part II, Metals and Materials, February 1983, pp. 90-96 and March 1973, pp. 145-151.

30. Welding Handbook, Vol. 1, Seventh Edition, American Welding Society, Miami,Florida, 1976.

31. Satoh, K. and Matsui, S., "Studies on Structural Restraint Severity Relating to Welding Cracking in Japan, IIW Doc. IX-986-76.

32. Lancaster, J. F., *Metallurgy of Welding*, published by George Allen and Unwin, London, 1980.

33. Zapffe, C. A. and Sims, C. E., "Hydrogen Embrittlement, Internal Stress and Defects in Steel," Trans. AIME, Vol. 145, 1941, p. 225.

34. Petch, N. J. and Stables, P., "Delayed Fracture of Metals Under Static Load," Nature, Vol. 169, 1952, p. 842.

35. Troiano, A. R., "The Role of Hydrogen and Other Interstitials in the Mechanical Behavior of Metals," Transactions of the ASM, Vol. 52, No. 1, 1960, pp. 54-80.

36. Oriani, R. A. and Josephic, W., "Testing of the Decohesion Theory of Hydrogen-Induced Crack Propagation," Scripta Met., 6, 1972, p. 681

37. Beachem, C. D., "A new Model for Hydrogen Assisted Cracking (Hydrogen Embrittlement)", Met. Trans., Vol. 3, February 1972, p. 437.

38. Savage, W. F., Nippes, E. F. and Tokunaga, Y., "Hydrogen Induced Cracking in HY-130 Steel Weldments," Welding Journal, Vol. 57, No. 4, April 1978, pp. 118s-126s.

39. Vasudevan, R., Stout, R. D. and Pense, A. W., "Hydrogen Assisted Cracking in HSLA Pipeline Steels," Welding Journal, Vol. 60, No. 9, September 1981, pp. 155s-168s.

40. Evans, G. M. and Baach, H., "Hydrogen Content of Welds Deposited by Different Welding Processes," Paper from Oerlikon Publication, "Hydrogen in Steel and Weld Metal," by G. M. Evans.

41. Keeler, T., "Inner-shield Welding," Metal Construction, Vol. 13, No. 11, November 1981, p. 667.

42. Specification for Carbon Steel Electrodes for Flux Cored Arc Welding--AWS A5.20-79, published by American Welding Society, Miami, Florida.

43. Schmid, G. C. and Rodabaugh, R. D., "A Water Displacement Method for Measuring Diffusible Hydrogen in Welds," Welding Journal, Vol. 59, No. 8, August 1980, pp. 217s-225s.

44. Nippes, E. F., Ball, D. J. and Gestal, Jr., W. J., "Determination of Diffusible Hydrogen in Weldments by the RPI Silicone Oil Extraction Method," Welding Journal, Vol. 60, No. 3, March 1981, p. 50s.

45. "The determination of the hydrogen content of ferritic arc weld metal," Welding in the World, Vol. 19, No. 5/6, 1981, pp. 57-67.

46. "Method for measurement of hydrogen evolved from deposited metal," Japanese Industrial Standard JIS Z 3113-1975.

47. Granjon, H., "The Implants Method for Studying the Weldability of High Strength Steels," Metal Construction and British Welding Journal, Vol. 1, No. 11, November 1969, p. 509.

48. Sawhill Jr., J. M., Dix, A. W. and Savage, W. F., "Modified Implant Test for Studying Delayed Cracking," Welding Journal, Vol. 53, No. 12, December 1974, pp. 554s-560s.

VITA

Srinivasarao Lathabai was born in Trivandrum, India on December 25, 1952. After passing the All India Higher Secondary Examination, conducted by the Central Board of Secondary Education, New Delhi, in the First Division in 1970, she entered the Indian Institute of Technology, Madras. She graduated in 1975 with a Bachelor of Technology in Metallurgy. She worked as a Development Engineer at the Welding Research Institute, Trichy, India from January 1976 to June 1980. Since August 1980, she has been a graduate student in the Department of Metallurgy and Materials Engineering at Lehigh University, Bethlehem, Pennsylvania.

She is a member of the American Society for Metals and the American Welding Society.