**NARROW GROOVE GAS TUNGSTEN ARC WELDING OF ASTM A508
CLASS 4 STEEL FOR IMPROVED TOUGHNESS PROPERTIES**

M. A. Penik Jr.

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NARROW GROOVE GAS TUNGSTEN ARC WELDING OF ASTM A508
CLASS 4 STEEL FOR IMPROVED TOUGHNESS PROPERTIES

by

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A Thesis Submitted to the Graduate
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CONTENTS

	PAGE
LIST OF TABLES.....	iv
LIST OF FIGURES.....	vi
ACKNOWLEDGEMENTS.....	ix
ABSTRACT.....	x
1.0 INTRODUCTION AND HISTORICAL REVIEW.....	1
1.1 Introduction and Objective.....	1
1.2 Historical Review.....	2
1.3 Narrow Groove Welding.....	15
2.0 RATIONALE' FOR THESIS.....	17
3.0 PROCEDURE.....	19
3.1 Weld Joint Design and Preparation.....	19
3.2 Welding Procedure.....	24
3.3 Specimen Preparation.....	30
3.3.1 Fracture Toughness Specimens.....	30
3.3.2 Charpy Impact Specimens.....	31
3.3.3 Tensile Specimens.....	32
3.3.4 Metallographic and Chemical Analysis.....	32
4.0 MATERIALS AND APPARATUS.....	39
4.1 Base Material.....	39
4.2 Weld Filler Material.....	43
4.3 Welding Equipment.....	49
5.0 TEST RESULTS.....	54
5.1 Weld Procedure.....	54
5.2 Fracture Toughness Test Results.....	59
5.3 Charpy Impact Test Results.....	64
5.4 Tensile Test Results.....	77

CONTENTS CONTINUED

	PAGE
5.5 Metallographic and Chemical Analyses.....	85
6.0 DISCUSSION AND CONCLUSIONS.....	91
6.1 Discussion.....	91
6.1.1 Weld Procedure and Effects on Metallurgy.....	91
6.1.2 Fracture Toughness	98
6.1.3 Charpy Impact.....	102
6.1.4 Tensile Tests.....	108
6.1.5 Metallographic and Chemical Analyses.....	113
6.2 Conclusions.....	120
7.0 LITERATURE CITED.....	122

LIST OF TABLES

	PAGE
Table 1.1 Comparison of Base and Weld Metal Charpy Impact Properties	14
Table 1.2 Typical Weld Metal Deposition Rates	14
Table 3.1 Narrow Groove Welding Procedure	29
Table 3.2 Test Matrix for Weld Wire Heats A and B	38
Table 4.1 ASTM A508 Class 4 Chemical Requirements	42
Table 4.2 ASTM A508 Class 4 Tensile Strength Requirements	42
Table 4.3 Filler Material Chemical Composition	48
Table 5.1 Final Narrow Groove Welding Procedure with Heat Input and Deposition Rate Calculations	58
Table 5.2 Heat A Fracture Toughness Test Results	62
Table 5.3 Heat B Fracture Toughness Test Results	63
Table 5.4 Charpy Impact Test Results for Heats A and B	76
Table 5.5 Heat A Tensile Properties	83
Table 5.6 Heat B Tensile Properties	84
Table 5.7 Heat A As-Deposited Chemical Analysis	89
Table 5.8 Heat B As-Deposited Chemical Analysis	90
Table 6.1 Fracture Toughness Data from Reference 6	101
Table 6.2 Charpy V-notch Test Data Comparison	107
Table 6.3 Tensile Data from Reference 6 and Heat 91H-005	112

LIST OF TABLES CONTINUED

	PAGE
Table 6.4 Chemistry and Mechanical Properties Comparison	119

LIST OF FIGURES

	PAGE
Figure 1.1 HSLA Steel Weldment Microstructure	11
Figure 1.2 HSLA Steel Weld Bead and HAZ	12
Figure 1.3 Iron-Carbon Phase Diagram with Overlapping HAZ	13
Figure 3.1 GTA Narrow Groove Joint Design	21
Figure 3.2 Mockup Design and Joint Fitup Requirements	22
Figure 3.3 Mockup Restrained for Welding	23
Figure 3.4 Mockups Used for Procedure Development	26
Figure 3.5 Microstructure during Weld Procedure Development	27
Figure 3.6 Bead Placement Variations in Narrow Groove Welds	28
Figure 3.7 Typical Fracture Toughness Specimen Layout	34
Figure 3.8 Typical Charpy Specimen Layout	35
Figure 3.9 Typical Tensile Specimen Layout	36
Figure 3.10 Cross Section for Metallography and Chemistry	37
Figure 4.1 Heat 91H-005 Base Material Microstructure	41
Figure 4.2 Model 215 and Hot Wire Power Supplies	51
Figure 4.3 Narrow Groove GTAW Torch Ready for Welding	52
Figure 4.4 Narrow Groove GTAW Torch Positioned in Weld Joint	53
Figure 5.1 Microstructure during Weld Procedure Development	56
Figure 5.2 Typical Weld Microstructure Using Final Weld Parameters	57

LIST OF FIGURES CONTINUED

	PAGE
Figure 5.3 Fracture Toughness Versus Temperature for Heats A and B	61
Figure 5.4 Charpy Impact Transition Curve for Heat A Weld 1	66
Figure 5.5 Charpy Impact Transition Curve for Heat A Weld 2	67
Figure 5.6 Charpy Impact Transition Curve for Heat A Weld 3	68
Figure 5.7 Charpy Impact Transition Curve for Heat B Weld 1	69
Figure 5.8 Charpy Impact Transition Curve for Heat B Weld 2	70
Figure 5.9 Charpy Impact Transition Curve for Heat B Weld 3	71
Figure 5.10 Charpy Impact Transition Curve Heat A Development Weld Fine-grained Region	72
Figure 5.11 Charpy Impact Transition Curve Heat A Development Weld Coarse-grained Region	73
Figure 5.12 Heat A Development Weld Showing Regions for Charpy V-notch Location	74
Figure 5.13 Typical Charpy V-notch in Refined Grain Region	75
Figure 5.14 Heat A Strength Versus Temperature	79
Figure 5.15 Heat B Strength Versus Temperature	80
Figure 5.16 Heat A and B Yield Strength Versus Temperature	81
Figure 5.17 Heat A and B Tensile Strength Versus Temperature	82
Figure 5.18 Heat A Narrow Groove Weld Metallography	87
Figure 5.19 Heat B Narrow Groove Weld Metallography	88

LIST OF FIGURES CONTINUED

	PAGE
Figure 6.1 Comparison of Thick and Thin Weld Beads	95
Figure 6.2 Thick Weld Beads and HAZ in Narrow Groove	96
Figure 6.3 Thin Weld Beads and HAZ in Narrow Groove	97
Figure 6.4 Fracture Toughness Comparison to Base Material	100
Figure 6.5 NDT ₃₅ ft.-lbs. Comparison	105
Figure 6.6 USE Comparison	106
Figure 6.7 Yield Strength Versus Temperature Comparison	110
Figure 6.8 Tensile Strength Versus Temperature Comparison	111
Figure 6.9 Typical Weld Microstructure from a Final Weld	116
Figure 6.10 Typical Microstructure from a Procedure Development Weld	117
Figure 6.11 Typical Microstructure from a Procedure Development Weld at High Magnification	118

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ABSTRACT

Welding of heavy section steel has traditionally used the automatic submerged arc welding (ASAW) process because of the high deposition rates achievable. However, the properties, particularly fracture toughness, of the weld are often inferior when compared to base material. This project evaluated the use of narrow groove gas tungsten arc welding (GTAW) to improve weld material properties. The welding procedures were developed for ASTM A508 Class 4 base material using a 1% Ni filler material complying to AWS Specification A.23-90-EF3-F3-N. A narrow groove joint preparation was used in conjunction with the GTAW process so competitive fabrication rates could be achieved when compared to the ASAW process.

Welding was performed with the cold- and hot-wire GTAW process using an ARC Machines Model 15 welding head. Welding procedures were developed for the flat position. The root pass and first 1/2" (12mm) of the weld joint were filled using the cold-wire GTAW process. Remaining fill was achieved using the split bead hot-wire GTAW process to increase the weld metal deposition rate.

Weld procedures were developed to refine weld substructure to achieve better mechanical properties. Two heats of weld wire were used to examine the effects of minor filler metal chemistry differences on weld mechanical properties. Extensive metallographic evaluations showed excellent weld quality with a refined microstructure. Chemical analysis of the weld metal showed minimal weld dilution

by the base metal. Mechanical testing included bend and tensile tests to ensure weld quality and strength. A Charpy impact energy curve versus temperature and fracture toughness curve versus temperature were developed for each weld wire heat. Results of fracture toughness and Charpy impact testing indicated an improved transition temperature closer to that of the base material properties.

1.0 INTRODUCTION AND HISTORICAL REVIEW

1.1 Introduction and Objective

American Society of Testing and Materials (ASTM) A508 Class 4 is a quenched and tempered high strength low alloy (HSLA) steel with attractive mechanical properties for pressure vessel applications (1,2). This steel offers high strength with exceptional fracture toughness (3). These characteristics allow the designer to make a strong, thin-walled vessel that is resistant to catastrophic brittle fracture allowing a through-crack to cause a leak and relieve pressure. ASTM A508 Class 4 has these qualities due to alloy additions and homogeneity of microstructure after heat treatment. Current manufacturing practices for joining ASTM A508 Class 4 forgings use the automatic submerged arc weld (ASAW) process. This process is preferred for joining thick sections due to an inherently higher deposition rate (4,5). Unfortunately, the full potential of ASTM A508 Class 4 has not been realized since the properties of ASAW deposited weld metal are inferior to those of the base steel (1).

The objective of this project was to investigate the potential for improving weld metal mechanical properties by switching to the gas tungsten arc weld (GTAW) process. A narrow groove weld joint was selected since it would make the GTAW process more competitive with the ASAW process in terms of fabrication rate.

1.2 Historical Review

Historically, the properties of steel weldments have been inferior to those of the steel base materials comprising the weldment (1). Previous testing of ASTM A508 Class 4 forgings and weld metal has consistently shown that the weld metal exhibits a lower resistance to brittle fracture than the base material in terms of Charpy V-notch impact testing (3,6,7).

Table 1.1 is a comparison of the available Charpy impact data for ASTM A508 Class 4 base metal and weld metal (deposited by various processes) from the open literature. Typical temperature values for the base material to exhibit 35 ft.-lbs. of energy range from -132°F to -87°F, while values for weld metal range from -65°F to -50°F. The large gap between weld and base material properties primarily exists because much of the research and development efforts have focused on forging fabrication techniques for HSLA steel that will enhance their resistance to aging, radiation embrittlement, and brittle fracture (1,6,8,9,10). Better processing techniques and forging methods, including heat treatments, have caused marked improvements in base material mechanical properties (11). New chemical techniques, used during ladle refining, reduce the tramp element content causing the steel to be cleaner; therefore, tougher, stronger, and more resistant to embrittlement aging (6,7,9,10).

There are technical and economic reasons for the weld metal not exhibiting

the same properties as the pressure vessel steel. Technically, the base material is homogenous and subjected to special heat treatments that improve toughness (8,11). Weld metals, on the other hand, are castings that are often expected to develop suitable properties in the as-stress relieved condition without the benefit of mechanical working or heat treatments that deform, refine, or completely change the cast microstructure and reduce any heterogeneity. Economically, suppliers of thick-section pressure vessels use welding processes that have higher deposition rates to achieve greater production rates (4,5,12,13,14).

Deposition rates for some of the typical weld processes used by manufacturers are shown in Table 1.2 (4,5). Fabricators in the United States and overseas generally select the ASAW process because of its much higher deposition rate which allows a wider, thicker weld bead to be deposited (15). Since this process is popular with most fabricators, much of the research in the literature has been directed towards improving the ASAW process.

Initial research evaluated the potential of increasing fabrication rates of the ASAW process by using a narrow groove joint design; however, this work failed to show any improvement in mechanical properties (14,16). More recent work has focused on improving mechanical properties of automatic submerged arc weldments. Fabricators realized they were unable to take full advantage of the superior base material; thus, the design of any vessel would be limited by the weld properties. Researchers geared much of their efforts for improving mechanical

properties of automatic submerged arc welds to developing new filler materials with different chemical composition. The critical elements that have the biggest impact on fracture toughness, strength, and ductility in welds are Mn, Ni, and Mo (1,12,17,18,19). However, for each element there is an optimal level above which there is no longer a noticeable improvement in mechanical properties. This recent work proved to be only partially successful, since the welds still did not have properties aligned with the base material (17,18,20,21). Also, research has revealed that the same elements that improve toughness have a detrimental effect on properties as the material thermally ages or is irradiated (9,10,22).

Research efforts have also been conducted on developing improved ASAW fluxes; since, ASAW flux composition affects mechanical properties (20,23). Typically, fluxes are designed to reduce the number of inclusions and oxides contained in the weld to improve the toughness (21,24,25,26,27). However, new flux formulations proved to be only partially successful in improving ASAW. The slight mechanical property improvements offered by new flux formulations are outweighed by the decrement induced by the solidification microstructure of an ASAW weld. ASAW inherently has a slow cooling rate, due to the higher heat input, which promotes the formation of a coarse grained solidification microstructure (17,18,28).

Most of the previous research has evaluated only chemistry changes to the filler metal or flux as they affect mechanical properties. But, chemical composition alone does not dictate properties. Weld mechanical properties are synergistically

determined by composition, microstructure (from cooling rate), and inclusion content (17,29).

The characteristics of the ASAW process, or any high deposition rate process, that make it attractive for heavy section steel welding are also what limit resulting mechanical properties. High deposition rate processes require greater heat input as compared to some other welding processes. This greater heat input results in a weld solidification substructure that is coarser; thus, resulting in reduced strength and toughness compared to finer substructures produced with less heat input (17,18,30,31). Minimal research efforts have been directed towards improving ASAW properties by substructure grain refinement because this would impact the high deposition rate. Dilution by the base material is also higher with the ASAW process because of the greater heat input. Base metal dilution affects the final composition of the weld metal and could potentially degrade mechanical properties.

The optimal microstructure in HSLA weldments is acicular ferrite (32,33,34). Acicular ferrite is characterized as an intragranularly nucleated phase of fine interlocked needles having high strength and toughness. It is the interlocking needles which give the weld metal extremely good fracture toughness, ductility, and strength. The formation of acicular ferrite is enhanced by fast weld cooling rates, low weld metal inclusion content, chemical composition, and weld metal substructure refinement due to subsequent weld passes (1,14,30). For example,

in terms of chemical composition, the effect of manganese in the 0.6% to 1.8% range was found to increasingly refine the weld microstructure and promote the formation of acicular ferrite (19).

When dealing with a transformation-hardening material, like HSLA steel welds, the most critical characteristic is the substructure refinement due to subsequent weld passes. Figure 1.1 shows a typical multi-pass HSLA steel weld. The last weld pass (top weld bead) is characterized by a coarse grain structure. During initial solidification the grains grow epitaxially from the fusion boundary. The size of the grains is determined by the weld solidification rate and temperature gradient (29). In HSLA steel welds, the initial solidification structure is typically coarse ferrite that nucleates on prior austenite grain boundaries (17,29,31). Ferrite forms upon cooling because the weld metal composition, primarily carbon content, is controlled (i.e., kept low) to prevent martensite formation (18,29). Grain boundary ferrite is known to have poor fracture toughness (1).

Figure 1.2 shows a schematic of an HSLA steel weld bead including the heat-affected zone (HAZ). The largest region is the fusion zone which was just discussed. The HAZ can be characterized by three regions: partial grain refined, grain refined, and grain coarsened regions (14,29). Figure 1.3 shows a schematic of an iron-carbon phase diagram. The partial grain refined region was subjected to a thermal cycle that peaked between the upper and lower critical temperature A_{c1} and A_{c3} . Most of the coarse grained ferrite in this region transformed to

austenite on heating above the Ac_1 and then decomposed into fine grained acicular ferrite upon cooling. However, some coarse grain boundary ferrite remained because the entire region was not completely austenitized. This region is therefore characterized by a mixed microstructure of coarse grain boundary ferrite and very fine grained acicular ferrite.

The grain refined region was subjected to a thermal cycle that peaked just above the Ac_3 , thus allowing the entire region to nucleate small austenite grains. Upon cooling, these small austenite grains completely decomposed into fine grained acicular ferrite in HSLA steel welds (29,33,34). Lastly, the grain coarsened region was subjected to temperatures that peaked above the Ac_3 , causing large austenite grains to form. Due to the large grains and high cooling rates, this region formed coarse blocky (polygonal) ferrite (18,35).

As Figure 1.2 displays, the HAZ contains both fine and coarse grained regions, but its average grain size is much smaller than that in the fusion zone (29). Therefore, with this knowledge of transformation-hardenable materials, weld beads can be strategically placed to cause nucleation of fine grained acicular ferrite throughout the weld metal fusion zone as shown in Figure 1.1. This microstructure will have improved mechanical properties.

As previously stated, acicular ferrite formation is enhanced by a fast weld cooling rate and low weld metal inclusion content, both which are characteristics of the GTAW process. GTAW provides a lower heat input which allows cooling at

much faster rates than ASAW. GTAW also uses an inert gas which assures a cleaner weld than ASAW; thus, reducing oxide inclusion content. These characteristics of the GTAW process result in a material that is inherently tougher. The smaller weld beads deposited with GTAW result in a finer grain and subgrain structure, and, as for all fusion processes, offers additional refinement during multipass welding. GTAW, however, is not used extensively for joining thick-section steel because it is more expensive to use than higher deposition rate processes. To make GTAW more competitive with ASAW in industry, a technique called narrow groove welding was developed (13).

Narrow groove welding not only offers a reduction in required weld metal, but also improves weld bead position control. Standard (non-narrow) groove weld joints require an increasing number of weld beads per layer as the weld joint is filled because of the large weld preparation angle. Weld beads placed in the joint cause areas of the fusion zone to contain fine and coarse grained regions. The mechanical properties of this type of weld are very different depending on which region is tested, because of the inhomogeneous microstructure (14,17,30).

In narrow groove welding, tight control is given to bead placement since there is a premium on weld joint space. The uniform positioning of weld beads, in turn, produces an even thermal history throughout the weld resulting in a more homogenous microstructure in the fusion zone (13,14). Therefore, the GTAW process and narrow groove welding technique work synergistically in producing a

homogeneously refined weld fusion zone.

There are other techniques that have been developed for refining weld metal microstructures (29,30). Researchers have found that coarse grained weld metal can be refined by dendritic fragmentation, grain detachment, or heterogenous nucleation (29,30). These three methods use special weld procedure techniques to induce refinement.

Dendritic fragmentation occurs in the weld metal mushy zone when tips of the growing dendrites break off and act as heterogenous nucleation sites for new grains. These new grains also block the large grains growing from the fusion boundary. Dendritic fragmentation can be produced by implementing weld pool stirring which influences the heat convection in the weld fusion zone (30).

Grain detachment occurs when partially melted grains become detached from the base material because of the liquid film that forms around them during welding (30). Like dendrite fragments, these grains act as heterogenous nucleation sites in the weld pool for new grains which refine the weld fusion zone. Grain detachment techniques have only been successful for alloys that exhibit a large partially-melted zone when using an arc oscillation technique.

Heterogenous nucleation is different than the two previous techniques. Heterogenous nucleation uses some type of inoculant as a source for the liquid atoms in the weld pool to build upon to form solidified grains. Inoculants are usually high melting temperature particles that are added to the weld pool. Typical

inoculants are Ti, W, or Zr particles. This technique has been successful in grain refining aluminum alloys (30), for example.

This project is unique in that it does not explicitly use any of these techniques to cause weld metal refinement. What is particularly unique in this project is it used the known characteristics of the GTAW process, the narrow groove welding technique, and the welding metallurgy of transformation-hardenable materials to improve weld metal properties. No special techniques were needed or employed.

Figure 1.1**HSLA Steel Weldment Microstructure**

Figure 1.2
HSLA Steel Weld Bead and HAZ

Columnar Weld Grains

Grain Coarsened Heat Affected Zone

Grain Refined Heat Affected Zone

Partial Grain Refined Heat Affected Zone

Figure 1.3

Iron-Carbon Phase Diagram with Overlapping HAZ

Table 1.1**Comparison of Base and Weld Metal Charpy Impact Properties**

Material	NDT ₃₅ (°F)
A508 HT 124S285 ³	-132
A508 HT 122Z195 ³	-87
A508 ID1 from NRL ⁷	-115
A508 ID2 from NRL ⁷	-110
ASAW HT 289246 ³	-65
ASAW HT 624116 ³	-63
ASAW HT IP2915 ⁷	-50

Note: NDT₃₅ is the temperature at which the material exhibits a minimum of 35 ft.-lbs. of energy as measured by Charpy impact testing.

Table 1.2**Typical Weld Metal Deposition Rates**

Process	Deposition Rate (lbs/hr)
Automatic Submerged Arc (ASAW)	15-30
Gas Metal Arc (GMAW)	6-18
Gas Tungsten Arc (GTAW)	3-6
Shielded Metal Arc (SMAW)	2-3

1.3 Narrow Groove Welding

The term "narrow groove welding" first appeared in the British Welding Journal in May 1966 (13). It has subsequently been used in the literature to describe a concept in welding in which the volume of weld metal required to complete a joint is significantly reduced relative to conventional practices; thus, saving labor, filler, time and cost.

An important point is that "narrow groove" does not describe a welding process. It refers to a special deposition technique along with the physical means of implementing it in a deep, narrow prepared joint. It is therefore possible to refer to any welding process when discussing narrow groove.

The extent to which the groove angle or the spacing of the faying surfaces can be reduced is limited by the physical dimensions of the equipment that must be inserted into the narrow groove for achieving adequate sidewall penetration. Techniques for "narrow groove" gas metal arc welding (GMAW) and ASAW have been developed (15,16,36,37). However, problems have been encountered with these two processes. Since they are high deposition processes, they cannot be used out of position and have a greater potential for causing lack of fusion defects in narrow grooves. ASAW also has the problem of slag detachment and new fluxes must be developed when used in narrow grooves(14).

The gas tungsten arc weld (GTAW) process in conjunction with a narrow groove joint design was selected for this application because it offers the following

advantages:

- * Reduced cost due to the narrow groove joint configuration.
- * Freedom from slag entrapment and oxide inclusions reducing toughness due to inert gas shielding.
- Flexible control of heat input, weld puddle size, and weld metal cooling rate.
- * Visual monitoring of the welding process.
- * Low heat input permitting grain refinement and tempering of the weld beads by subsequent passes.
- * Ability to make easy repairs with the same process, whether using the automatic or manual mode.
- * High weld quality.

The goal of this study was to develop a GTAW narrow groove process for ASTM A508 Class 4 material that would exhibit weld metal properties comparable to the base material. This would allow for a more balanced pressure vessel design. This thesis submits the results of testing narrow groove gas tungsten arc welds including mechanical properties, chemistry, and metallographic evaluations.

2.0 RATIONALE' FOR THESIS

As discussed earlier, the GTAW process allows for better heat input control and bead placement within the weld joint than conventional high deposition rate welding processes. The smaller, thinner weld beads result in a faster cooling rate causing a refined acicular ferrite microstructure to form. The premise tested in this thesis was whether a weld procedure could be developed that consistently produced a refined microstructure while filling a narrow groove joint design at relatively high deposition rates using the hot-wire gas tungsten arc welding (HW-GTAW) process. The refined microstructure was necessary to achieve mechanical properties, particularly, good toughness. The narrow groove joint design was selected to offset the inherently slower deposition rate of the GTAW process. Also, quality welds free from defects were mandatory, since defects would adversely affect performance of pressure vessels.

A secondary premise tested in this thesis was whether slight variations in weld wire chemistry affect properties. It was theorized that not only the refined microstructure; but potentially small variations in Ni, C, and Mo content may also affect toughness. To test this theory, two weld wire heats procured to AWS Specification A5.23-90 (38), were examined using the same welding procedure. One heat of ASTM A508 Class 4 material was used during all procedure development and test welding so that base material chemistry would not influence the testing results.

Complete characterization of the welds was performed to test the theories. Tensile, Charpy impact, and J_{IC} fracture toughness tests were used to measure the mechanical properties of the welds. Metallography permitted characterization of the microstructure, and verified it was refined throughout the weld fusion zone. Chemical analysis was used to determine the chemical composition of each weld, and determine how it influenced weld mechanical properties.

3.0 PROCEDURE

3.1 Weld Joint Design and Preparation

Figure 3.1 details the gas tungsten arc (GTA) narrow groove weld joint design. An 8° included angle was used because it was the smallest angle permissible with the weld torch used. A 0.250" root radius with a nominal 0.095" root land was selected since these offered good weld joint fitup and access for root bead welding. The weld joint was nominally 3" thick.

Each weld mockup was fabricated from two pieces of material saw cut from the same large forging prolongation. Both pieces were then machined to include the narrow groove joint design using standard machining techniques. Mockup pieces were then cleaned using hot water and detergent to remove cutting fluid and any other foreign material. After cleaning the mockups were placed under heavy clean cloth to protect the material from the environment prior to welding.

A typical weld mockup is shown in Figure 3.2. The mockups were designed to be prototypical of a weld heat sink seen in actual applications. The weld was selected to be 3 inches thick by 24 inches long to allow for sufficient mechanical test specimens. Root gap fitup was critical, and was controlled to the requirements of Figure 3.2 so that a smooth weld underbead was obtained. Each mockup was fully restrained by fillet welding to a carbon steel backing plate as shown in Figure 3.3. Restraining the mockup also better represented typical conditions of pressure vessel fabrication in which there is significant restraint as fabrication (welding)

progresses. All eight mockups were fabricated from the same heat of ASTM A508 Class 4 material.

Figure 3.1

GTA Narrow Groove Joint Design

Figure 3.2

Mockup Design and Joint Fitup Requirements

DETAIL A

Figure 3.3

Mockup Restrained for Welding

3.2 Welding Procedure

Initial parameter development was completed on small mockups such as shown in Figure 3.4. The smaller mockups were cheaper to fabricate and easier to handle while still representing prototypical conditions. The objective of parameter development was to achieve grain refinement throughout the entire weld fusion zone. Figure 3.5 reveals the microstructure obtained during early parameter development. Coarse grains were found along the weld fusion zone while refined acicular ferrite was dominate along the center of the weld. Further parameter development was necessary to achieve complete weld metal refinement.

Table 3.1 lists the final weld procedure used for all eight test mockups. These parameters were successful in refining the weld microstructure for both heats of weld wire. Except for the root pass, two beads were deposited for every weld layer. Figure 3.6 shows some typical variations in weld bead placement for narrow groove weldments. For this project, the two bead per layer technique was selected. This technique is often called "split bead" since it requires two weld beads of equal size to complete a layer.

The root pass and first 1/2" (12mm) of weld metal was deposited with the cold wire GTAW process. This created a smooth root bead profile and substantial weld buildup for the remaining (hotter) fill passes. Argon shielding gas was used through the torch to protect the weld puddle and arc. The root side of the weld joint was purged with argon to prevent oxidation. Once the first 1/2" (12mm) was

deposited, the weld joint was wide enough for depositing hot wire GTA weld beads. A mixture of 75% helium and 25% argon shield gas was used during hot wire GTA welding. The mixed shield gas created a hotter welding arc, ensuring weld bead tie-in. As Table 3.1 shows, greater wire feed rates were achieved as welding progressed since the weld joint is larger near the top; thus, allowing larger beads to be deposited.

Prior to welding, each mockup was preheated to 200°F to prevent hydrogen cracking by precluding the formation of untempered martensite from rapid cooling above the critical cooling rate. An interpass temperature of 350°F was selected and measured after every weld bead. Once the mockup reached 350°F, no further welding was permitted until it cooled to the preheat temperature. The narrow 150°F difference between preheat and interpass was critical since the formation of acicular ferrite is dependent upon the weld cooling rate. After welding, each mockup received a post weld stress relief treatment for 30 hours at 1050°F. This heat treatment is typical of industry (1,3,8,16).

Figure 3.4
Mockups Used for Procedure Development

Figure 3.5**Microstructure during Weld Procedure Development**

Figure 3.6

Bead Placement Variations in Narrow Groove Welds

Variations in narrow groove bead placement

- A: Single bead per layer
- B: Two beads per layer/ Split Bead
- C: Three beads per layer

Table 3.1

Narrow Groove Welding Procedure

Parameter	Root and cold wire	First 1/2" hot wire	Next 1/2" hot wire	Next 1" hot wire	Final 1/2" hot wire
Avg amps	150-200	320	320	320	320
Pulse Hi (sec)	.30	-	-	-	-
Pulse Low (sec)	.30	-	-	-	-
Wire feed (ipm)	40-60	125-150	150-175	175-185	185-200
Volts	10-12	10-13	10-13	10-13	10-13
Torch tilt (degrees)	0-7	0-7	0-7	0-7	0-7
Osc (in)	-	1/16	1/16	1/16	1/16
Osc freq (cyc/min)	-	105/115	105/115	105/115	105/115
Hot wire Amps	-	100-115	100-120	100-120	100-120
Hot wire Volts	-	2-4	2-4	2-4	2-4
Tvl speed (ipm)	3-5	6.5	6.5	6.5	6.5
Wire size (in)	0.035	0.035	0.035	0.035	0.035
Shield gas (CFH)	80	80	80	80	80

Legend:

Avg amps = Average welding current in amperes

Pulse = Duration the current was at current level measured in seconds

3.3 Specimen Preparation

Four welds were made with each heat of weld wire. The first four welds were made with Heat A filler wire, and were identified as A1, A2, A3, and A4 for traceability. Identification of the second set of four welds, which used Heat B filler wire, was B1, B2, B3, and B4. Enough weld material was deposited to ensure sufficient specimens could be removed for testing. After stress relief, the welds were machined to extract and produce mechanical test specimens, and chemistry and metallography samples. All mechanical test specimens were machined, inspected, and tested to the requirements of applicable ASTM specifications. Table 3.2 details the test matrix for each heat of weld wire.

3.3.1 Fracture Toughness Specimens

Seventeen 1.3T (1.3" thick) compact tension (CT) specimens were machined from welds made with each heat of weld wire as shown in Table 3.2. Specimens were tested at various temperatures to develop a fracture toughness curve for a narrow groove GTAW. All specimens were machined and tested in accordance with ASTM E813-89 specifications (39) on a computer controlled 55 Kip machine.

A typical weld layout for removing and machining fracture toughness specimens is shown in Figure 3.7. Two specimens were removed through the thickness of the weld, centered around the mid-thickness of the weld. The first and last inch of the weld coupon were not used since they contained weld starts and

stops. The crack propagation direction for all specimens was parallel with the welding direction, since previous results from the literature found this to be the most limiting condition (17,40).

3.3.2 Charpy Impact Specimens

Twenty-four Charpy V-notch (CVN) specimens are required to produce a full ductile-to-brittle transition curve of impact energy versus test temperature (41). Three specimens were tested at each of eight different temperatures for consistency in creating a meaningful transition curve. Three full transition curves were generated for each heat of weld wire using data on each test from Table 3.2. This was done to determine if there was a significant difference between weld wire heats, or different welds made with the same heat of wire.

From each curve, the most critical data collected are the upper shelf energy and the temperature at which 35 ft.-lbs. of energy is exhibited, also called the "nil-ductility temperature at 35 ft.-lbs." (NDT_{35}). The upper shelf energy (USE) is a measure of toughness. The higher the USE, the tougher the material. The nil-ductility temperature is an indicator of the material's resistance to brittle fracture. Lower values of NDT_{35} indicate that the material is ductile to lower temperatures.

All Charpy specimens were machined and tested in accordance with ASTM E23-94b specifications (41) on a Tinius Olsen 400 ft.-lbs. pendulum-type machine. A typical weld layout for removing and machining Charpy specimens is shown in Figure 3.8. Four specimens were removed through the thickness of the weld,

centered around the mid-thickness of the weld. The crack propagation direction for all specimens was parallel to the welding direction to be consistent with the fracture toughness specimens.

3.3.3 Tensile Specimens

Twelve tensile specimens were tested from welds made with each heat of weld wire as shown in Table 3.2. Temperatures for testing were selected based on those used during fracture toughness testing, since yield strength and Young's modulus are needed to validate fracture toughness properties (39). Specimens were removed from the welds as shown in Figure 3.9. The tensile specimens were machined parallel to the weld so they would contain all weld metal and thus provide a true measure of the narrow groove weld tensile properties. All tensile specimens were machined and tested in accordance with ASTM E8-92 and E21-92 specifications (42,43) on an Instron 4510 Test Station. Yield strength, tensile strength, % reduction area, and % elongation were measured and recorded for each test.

3.3.4 Metallographic and Chemical Analysis

Cross sections from welds made with each weld wire heat were prepared for metallography by polishing and etching with a 2% nital solution. Samples were examined with the optical microscope to reveal microstructural constituents. References (33,34,35) were used to identify and quantify the different phases present in the weld metal. After metallography, the cross section was cut as shown

in Figure 3.10 for chemical analysis. Pieces were removed from the bottom, middle, and top of the weld to quantify the actual weld metal chemistry. This analysis was used to identify the amount of dilution by base material and to determine the effect weld metal chemistry has on mechanical properties.

Figure 3.7
Typical Fracture Toughness Specimen Layout

Figure 3.8**Typical Charpy Specimen Layout**

Typical layout of Charpy specimens in weld mockup

Figure 3.9

Typical Tensile Specimen Layout

Figure 3.10

Cross Section for Metallography and Chemistry

Table 3.2

Test Matrix for Weld Wire Heats A and B

Test type	Temperature							
	T _{LS1}	T _{LS2}	T _{T1}	T _{T2}	T _{T3}	T _{T4}	T _{US1}	T _{US2}
Fracture Toughness	2	2	3	3	3	-	2	2
CVN-Curve1	3	3	3	3	3	3	3	3
CVN-Curve2	3	3	3	3	3	3	3	3
CVN-Curve3	3	3	3	3	3	3	3	3
Tensiles	All-weld-metal, 12 tensiles done at various temperatures to support fracture toughness.							
Chemistry	Metallography and chemistry done on typical weld cross sections							

Notes:

T_{LS} Lower shelf temperature
 T_T Transition regime temperature
 T_{US} Upper shelf temperature

4.0 MATERIALS AND APPARATUS

4.1 Base Material

One heat of ASTM A508 Class 4 material was used for all mockups, thereby eliminating any possible effects due to base material variations. The material was supplied as a single forging prolongation that was identified as Heat 91H-005.

ASTM A508 Class 4 is a high strength low alloy quenched and tempered steel that exhibits high strength and toughness (3,6,7). The ASTM specification for this material requires that after forging to final configuration the material first be austenitized in the temperature range 1540°F to 1640°F for 1 hour per inch of thickness. After austenitizing, the material is quickly quenched by immersion in a tank of brine solution to produce a mixed martensitic and bainitic microstructure throughout the forging (2). Tempering of the forging is then initiated at a subcritical temperature of 1125°F for a minimum time of 1/2 hour per inch of thickness. The final microstructure is very uniform, consisting of tempered martensite and bainite. It is the mechanical working, heat treatment and chemical composition that makes this steel so tough and strong (2,6,7,9). Materials like ASTM A508 Class 4 are usually used for pressure retaining vessels because they are very pure, homogenous and exhibit isotropic properties (6).

Table 4.1 details the chemical requirements for ASTM A508 Class 4 and also lists the chemistry of Heat 91H-005. ASTM A508 Class 4 is a nominal 3% Ni

steel. Nickel and molybdenum are the two key elements that give this material such good toughness (6,7,9). Tramp elements such as sulfur and phosphorus are tightly controlled since they are detrimental to toughness (9,10). Heat 91H-005 met the chemical requirements of the specification. Note in particular this heat has very low levels of tramp elements.

The specification for ASTM A508 Class 4 also requires base material tensile testing at room temperature (75°F). Table 4.2 shows the minimum requirements set forth in the specification, and the tensile results for Heat 91H-005. Three specimens were tested at room temperature to determine an average strength value for meeting the specification requirements. The results from the base material test not only showed that the material meets the specification requirements, but also indicated that the material is very homogenous since the values from all three tests were very similar.

Figure 4.1 reveals the microstructure of Heat 91H-005. Photomicrographs at 50 times and 425 times magnification are shown. The microstructure has a very uniform grain size. Tempered martensite and bainite are the predominant structures that characterize this material, which is consistent with the specification requirements.

Figure 4.1

Heat 91H-005 Base Material Microstructure

50x magnification

425x magnification

Table 4.1**ASTM A508 Class 4 Chemical Requirements**

Element % Weight	ASTM A508 Class 4 Specification	Heat 91H-005
C	0.23 max	0.203
Mn	0.20-0.40	0.25
P	0.020 max	0.003
S	0.020 max	0.008
Si	0.10 max	0.03
Ni	2.8-3.9	3.33
Cr	1.50-2.00	1.66
Mo	0.40-0.60	0.46
Cu	0.20 max	0.09

Table 4.2**ASTM A508 Class 4 Tensile Strength Requirements**

	Test Temp (F°)	0.2% YS (Ksi)	UTS (Ksi)	Elongation (%)	Reduction Area (%)
ASTM- A508 Cl 4	75	85 min	105-130	18	45
91H-005-1	75	99.5	117.4	25	71.9
91H-005-2	75	99.5	115.9	25	74.2
91H-005-3	75	98.6	115.0	26	74.9

4.2 Weld Filler Material

Two heats of filler material were procured to ANSI/AWS A5.23-90 (38) specification. This specification prescribes requirements for the classification, chemistry, and quality of low alloy steel electrodes for welding. Most of the welding electrodes in this specification are used for ASAW, however they can be used with the GTAW process as well. Industry uses American Welding Society (AWS) specifications when ordering filler material to ensure the manufacturer certifies that the product meets the minimum specification requirements.

The particular classification for the welding electrode used in this project was AWS A5.23-90-EF3-F3-N. This classification details the chemical limits of the filler material as shown in Table 4.3. Thyssen Specialty Steels (44) was the manufacturer of the filler material. Two Heats, Heat A and Heat B, were purchased in a 0.035" diameter wire size suitable for GTA narrow groove welding. Thyssen was selected since they are a worldwide producer of quality welding products. The chemical composition of each heat is also shown in Table 4.3.

Chemical composition is one of the factors which will influence as-deposited weld metal mechanical properties. The general trends of individual alloying elements on weld metal mechanical properties are discussed in this section of the thesis. The same trends would be expected for the base material.

Carbon

Carbon is added to filler materials in low to moderate levels to enhance

strength, but causes ductility to decrease (45). Carbon increases the Charpy V-notch transition temperature and lowers the upper shelf energy (USE). Carbon is usually controlled in filler materials to below 0.20% to prevent martensite from forming during solidification (1). Notice that the AWS specification, Table 4.3, tightly controls carbon within a 0.10-0.18 weight percent band, while the ASTM A508 Class 4 base material specification, Table 4.1, allows a more liberal 0.23 weight percent maximum.

Manganese

At the strength levels of welds used to join most pressure vessels, manganese usually increases the strength and toughness of weld metals up to 1.6 weight percent (19). Manganese increases the hardenability of weld metals and tends to promote acicular ferrite which improves toughness (19). It is used as a strengthener since carbon levels are lower in weld metals. Manganese also deoxidizes the weld pool, reducing the number of weld inclusions, and tying up sulfur to aid in preventing solidification cracking. Both Heat A and Heat B had an optimal amount of manganese.

Silicon

The primary function of silicon is for deoxidation of the weld pool. Silicon combines with various elements to form harmless silicates and prevent porosity (1). Additions of silicon up to 0.35 weight percent also aid in weld puddle fluidity for most alloy steels. At this level, silicon is not a major concern for affecting

toughness. Base materials have much lower silicon levels since it is not needed for the purposes described.

Sulfur/ Phosphorus

Both sulfur and phosphorus are residual (tramp) elements resulting from impurities during fabrication; consequently, having a harmful effect on weldability and toughness (1,29). Additionally, these elements are known to adversely affect properties as the weld and base material are exposed to elevated temperature or radiation for long periods of time (7,10). Thyssen materials are known for their low levels of tramp elements. Both Heats A and B were well below the maximum value allowed for these elements. It is also worth noting that the base material had very low impurity levels.

Chromium

Chromium influences strength because it forms carbides which harden the weld metal; thus, it also plays an important role in influencing weld metal toughness. For HSLA weld metals, chromium is a moderate strengthener, and can be added up to 0.5 weight percent without seriously impairing toughness. Changes in toughness with chromium depend upon other elements present in the weld metal (1). However, for this particular filler material it is considered a tramp element, and is not used purposely to strengthen or toughen the weld metal. Thyssen has been able to control chromium content to 0.10 weight percent maximum for the heats tested. Notice, however, that in the base material the

chromium level is 1.66 weight percent and is used to form carbides during quenching, which harden the base material matrix. Chromium in this case is used as an alloy element for strengthening and increasing toughness.

Nickel

Of all the alloy elements present in this filler material, nickel has the most effect on toughness (1,9). Nickel lowers the Charpy V-notch transition temperature when added in amounts up to 1.5 weight percent in stress relieved welds. Above this level, nickel enhances temper-embrittlement and radiation-embrittlement of welds (1,3,9,10). The AWS specification allows a wide range for nickel composition (0.70%-1.10%), Table 4.3, in this filler material. The Thyssen heats both fall on the upper end of this limit and; therefore, would have better toughness than heats with lower nickel, while still maintaining resistance to embrittlement. Notice that the base material nickel level is much different (2.8%-3.9%), Table 4.1, than the filler material. This is the major reason why ASTM A508 Class 4 forgings are tougher than the welds that join them.

Molybdenum

Molybdenum is a reasonably potent strengthening agent in steel weld and base metals (1). Generally, molybdenum is added in amounts from 0.20-0.60 weight percent for strengthening purposes (1). The advantage of molybdenum lies in its ability to strengthen metal without a significant reduction in toughness. Both the filler material and base material used in this project have comparable amounts

of molybdenum for this purpose.

Copper

Copper has a mild strengthening effect on HSLA steel welds. When welds contain less than 0.5 weight percent copper, the copper does not have a strong influence on Charpy V-notch transition temperature. Copper is also known to embrittle HSLA welds and forgings when used for nuclear applications (3,7,9,10,22,46). Therefore, the AWS and ASTM specifications put a very tight control on copper content.

Table 4.3**Filler Material Chemical Composition**

Element % Weight	AWS Spec.	Heat A	Heat B
C	0.10-0.18	0.14	0.11
Mn	1.25-2.25	1.60	1.66
Si	0.20 max	0.14	0.16
S	0.025 max	0.005	0.005
P	0.025 max	0.008	0.005
Cr	-	0.10	0.05
Ni	0.70-1.10	0.90	0.97
Mo	0.40-0.65	0.53	.54
Cu	0.35 max	0.19	0.02
Fe	Rem	Rem	Rem

4.3 Welding Equipment

All welding was performed with an ARC Machines Incorporated (AMI) Model 215 power supply and Model 15 pipe weld head adapted with a narrow groove GTAW torch. The Model 215 power supply is shown in Figure 4.2. Also shown in the figure is the power supply used for supplying current to the wire guide nozzle during HW-GTAW. Figure 4.3 shows the narrow groove torch adapted to the Model 15 pipe weld head which has both hot and cold wire GTAW capabilities. Figure 4.4 shows the narrow groove torch positioned in the weld joint. The narrow groove torch is specifically designed for use with the 8° weld joint design.

The Model 215 power supply is microprocessor controlled. All welding parameters are contained within the power supply in specific files, very similar to computer files. When welding is initiated, the weld operator calls up the file for use and presses the welding sequence start button. The machine then takes over by initiating the welding arc and stabilizing the essential welding parameters contained in the file. This type of equipment is called "synergic", and is highly automated, so the operator only needs to monitor torch position within the weld joint.

One of the unique features of the Model 215 power supply is that it has a lockout device which prevents the operator from modifying welding parameters out of specific tolerances. This is essential for repetitive welding where mechanical properties are dependent upon the welding parameters selected. Once procedure parameters are developed, they cannot be modified since the changes would have

an unknown effect on mechanical properties. Use of this equipment was essential on this project since welding parameters play such a significant role in determining mechanical properties for HSLA steel welds (1). Additionally, it ensured that all eight test mockups were welded using the same parameters so a direct comparison could be made between filler metal Heat A and Heat B.

Figure 4.2
Model 215 and Hot Wire Power Supplies

Figure 4.3

Narrow Groove GTAW Torch Ready for Welding

Figure 4.4

Narrow Groove GTAW Torch Positioned in Weld Joint

5.0 Test Results

5.1 Weld Procedure

The welding procedure was developed to be simple, easy to use, and result in complete weld metal grain refinement in the fusion zone. Initial development efforts focused on maintaining welding parameters the same throughout the weld joint. For example, initial development used a wire feed speed of 200 ipm throughout the weld. However, as shown in Figure 5.1, the weld metal microstructure showed regions of coarse grains along the fusion zone. Table 5.1 shows the final weld procedure parameters used for all test weldments. Only the parameters which affect heat input and deposition rate are shown, for a complete listing of the parameters see Table 3.1.

The welding parameters were kept as simple as possible. The root and cold wire portion of the weld joint was completed with the same parameters. During hot wire welding, the only parameter that was varied was wire feed speed. This parameter seemed to have the greatest influence on the weld microstructure and will be discussed in greater detail later. The wire feed speed was gradually increased as the joint was filled since the weld groove becomes larger and could tolerate a bigger weld bead.

Table 5.1 also shows heat inputs and weld metal deposition rates calculated using the equations given at the bottom of the table. Hot wire welding was performed with the same heat input throughout the weld joint. The deposition

rate gradually increased since the wire feed speed increased as the weld joint was filled. Both the cold- and hot-wire welding heat inputs were kept low to enhance the weld cooling rate used to promote acicular ferrite formation (33). Figure 5.2 shows the weld fusion zone of a narrow groove weld made with Heat A using the weld parameters in Table 5.1. Notice that the weld metal is now fully refined when compared to the initial microstructure in Figure 5.1.

Figure 5.1

Microstructure during Weld Procedure Development

Figure 5.2

Typical Weld Microstructure Using Final Weld Parameters

Table 5.1

**Final Narrow Groove Welding Procedure with
Heat Input and Deposition Rate Calculations**

Parameter	Root and cold wire	First 1/2" hot wire	Next 1/2" hot wire	Next 1" hot wire	Final 1/2" hot wire
Avg amps	150-200	320	320	320	320
Wire feed (ipm)	40-60	125-150	150-175	175-185	185-200
Volts	10-12	10-13	10-13	10-13	10-13
Tvl speed (ipm)	3-5	6.5	6.5	6.5	6.5
Wire size (in)	0.035	0.035	0.035	0.035	0.035
Shield gas (CFH)	80	80	80	80	80
Heat Input (KJ/in)	28.9	34	34	34	34
Deposition Rate (lbs/hr)	0.84	2.3	2.7	3.0	3.2

Heat Input (KJ/in) = (Volts × Amps × 60)/Tvl speed

Deposition Rate (lbs/hr) = $\rho \times A \times ((B)^2 \times \pi / 4) \times 60 \text{ min/hr}$

Density = ρ = 0.290 lbs/in³ for HSLA weld metal (44)

Wire Feed (ipm) = A

Wire size (in) = B

5.2 Fracture Toughness Test Results

Seventeen 1.3 CT specimens were tested from welds made with each heat of weld wire were tested. Tables 5.2 and 5.3 list the results, and Figure 5.3 plots the results of all fracture toughness testing. Also drawn in Figure 5.3 is a conservative fracture toughness curve. J_{Ic} values were converted to K_{Ic} values using the equation shown at the bottom of the table. K_{Ic} values were then used for comparison to data in the literature. Testing was conducted over the temperature range of -290°F to 500°F to construct a fracture toughness curve. Most specimens were tested at low temperatures since this establishes the transition regime of the curve, and is the most difficult region for obtaining valid results. Specimens from each heat were tested predominately at the same temperatures for comparison purposes. However, some additional temperatures were used so the fracture toughness curve could be completely filled.

Specimen A6 from Heat A and specimen B4 from Heat B are not plotted in Figure 5.3. Both of these specimens were tested at 500°F and used the same special high temperature clip gage. Unfortunately, this gage did not perform well during testing and caused the tests to become invalid to ASTM requirements. Further testing at 500°F used a different gage.

Notice from Figure 5.3, and the conservative fracture toughness curve, that the transition regime ranges from approximately -260°F to -150°F. In this regime, the material behaves in a more brittle than ductile manner. Above this temperature,

the material is predominantly on the upper shelf and behaves in a ductile manner. The nominal fracture toughness value on the upper shelf is $200 \text{ Ksi}\sqrt{\text{in}}$, based on the fracture toughness curve. At all the similar temperatures tested, Heat B had better fracture toughness than Heat A, except at room temperature (75°F). Also notice there is scatter in the data for specimens tested at the same temperature for each heat. This is attributed to the complex nature of performing fracture toughness testing, and confirms the necessity for running duplicate specimens to establish nominal fracture toughness values.

Figure 5.3

Fracture Toughness Versus Temperature for Heats A and B

Table 5.2

Heat A Fracture Toughness Test Results

Specimen	Temperature (°F)	J _{IC} (in-lbs/in ²)	K _{IC} (Ksi√in)	Modulus (E ⁶)
A1	75	1546.0	226.5	30.2
A2	75	1219.0	201.1	30.2
A3	0	1335.0	211.9	30.6
A4	0	1377.0	215.2	30.6
A5	500	1060.0	180.3	27.9
A6 ¹	500	347.9	103.3	27.9
A7	-150	996.0	185.4	31.4
A8	-165	1300.0	212.1	31.5
A9	-175	471.6	127.9	31.6
A10	-175	579.3	141.7	31.6
A11	-200	556.3	139.9	32.0
A12	-200	543.1	137.5	32.0
A13	-225	165.9	76.2	31.8
A14	-225	133.5	68.3	31.8
A15	-250	55.7	44.3	32.0
A16	-250	29.6	32.3	32.0
A17	-261	21.0	27.3	32.1

1: Specimen A6 was invalid due to difficulties with a test clip gage.

$$K_{IC} = \sqrt{((J_{IC}E)/(1-\nu^2))}$$

$\nu = 0.3$ (Poisson's ratio)

Table 5.3

Heat B Fracture Toughness Test Results

Specimen	Temperature (°F)	J_{IC} (in-lbs/in ²)	K_{IC} (Ksi√in)	Modulus (E ⁶)
B1	75	874.0	167.7	30.2
B2	75	1229.0	201.9	30.2
B3	0	1506.0	246.3	30.6
B4 ¹	500	168.0	22.7	27.9
B5	500	2119.0	254.0	27.9
B6	500	2043.0	250.0	27.9
B7	-175	862.4	172.9	31.6
B8	-200	181.3	79.5	31.7
B9	-200	416.1	120.4	31.7
B10	-200	594.2	143.8	31.7
B11	-225	395.8	117.6	31.8
B12	-225	337.1	108.6	31.8
B13	-250	344.3	109.9	31.9
B14	-250	539.4	137.6	31.9
B15	-250	103.3	60.2	31.9
B16	-250	275.9	98.5	31.9
B17	-290	98.8	56.1	29.0

1: Specimen B4 was invalid due to difficulties with a test clip gage.

$$K_{IC} = \sqrt{((J_{IC}E)/(1-\nu^2))}$$

$\nu = 0.3$ (Poisson's ratio)

5.3 Charpy Impact Test Results

NDT₃₅ ft.-lbs. and USE values from welds made with Heats A and B are listed in Table 5.4. Figures 5.4, 5.5, and 5.6 show the full transition curves for Heat A welds. Curves for Heat B welds are shown in Figures 5.7, 5.8, and 5.9. Three full Charpy impact transition curves were generated for each heat of weld wire using the parameters shown in Table 5.1. Twenty-four specimens were needed to make each curve. From each curve the NDT₃₅ ft.-lbs. and USE was collected. Three curves were generated to establish a nominal NDT₃₅ ft.-lbs. value for each weld wire heat. The Charpy impact energy values from each data set were input into a hyperbolic plotting subroutine to form fit the data into a transition curve. This subroutine was used to maintain consistency across all data sets so the data could be compared to establish the effect weld wire chemistry had on toughness.

Also shown in Table 5.4 are NDT₃₅ ft.-lbs. and USE values from initial development welds made with weld wire Heat A. The curves for these data are shown in Figures 5.10 and 5.11. Charpy V-notch specimens were taken from the fine-grained and coarse-grained regions of the development mockup as shown in Figure 5.12. The V-notch was centered on the line, as shown in the figure, to characterize each region. A complete Charpy curve was generated so an NDT₃₅ ft.-lbs. and USE could be determined for each region.

Figure 5.13 shows a typical Charpy specimen taken from the center of the weld joint where the weld metal was completely refined. Notice that on either side

of the notch there are coarse grains. For the coarse-grained Charpy specimens, the notch would obviously be moved to the right. Reviewing the data shows the grain refined region exhibits an NDT_{35} ft.-lbs. of -99°F , which is better than the coarse-grained region value of -77°F . Notice, however, that the USE for both regions is very similar. It appears that microstructure has a more pronounced affect on transition temperature than on USE.

Table 5.4 reveals that Heat B consistently out-performed Heat A in both NDT_{35} ft.-lbs. and USE. The average values for each heat are shown in Table 5.4. Heat B was nominally better by approximately 50°F in NDT_{35} ft.-lbs., and 22 ft.-lbs. in USE.

Figure 5.4

Charpy Impact Transition Curve for Heat A Weld 1

Figure 5.5

Charpy Impact Transition Curve for Heat A Weld 2

Figure 5.6

Charpy Impact Transition Curve for Heat A Weld 3

Figure 5.7

Charpy Impact Transition Curve for Heat B Weld 1

Figure 5.8

Charpy Impact Transition Curve Heat B Weld 2

Figure 5.9

Charpy Impact transition Curve for Heat B Weld 3

Figure 5.10

Charpy Impact Transition Curve for Heat A Development Weld
Fine-grained Region

Figure 5.11

**Charpy Impact Transition Curve for Heat A Development Weld
Coarse-grained Region**

Figure 5.12

Heat A Development Weld Showing Regions for Charpy V-notch Location

Figure 5.13**Typical Charpy V-notch in Refined Grain Region**

Table 5.4

Charpy Impact Test Results for Heats A and B

Heat A			
Figure	Weld	NDT ₃₅ (°F)	USE (ft-lbs)
5.10	Development Refined Grains	-99	137.6
5.11	Development Coarse Grains	-77	147.0
5.4	1A	-129	136.1
5.5	2A	-124	142.1
5.6	3A	-146	135.1
Average		-133	137.7
Heat B			
5.7	1B	-172	169.9
5.8	2B	-175	156.2
5.9	3B	-204	155.6
Average		-184	160.6

5.4 Tensile Test Results

To ensure the weld exhibited sufficient strength, and for validation of the fracture toughness data, it was necessary to conduct a tensile testing survey for each heat over the temperature range of -320°F to 650°F. The results of tensile testing for Heats A and B are given in Tables 5.5 and 5.6, respectively. For comparison, the ASTM A508 Class 4 specification and AWS specification room temperature requirements are also listed in the tables. The yield and tensile strength for Heats A and B are plotted in Figures 5.14 and 5.15, respectively.

For both heats, yield and tensile strength decreased significantly with increasing test temperature. However, above 0°F the decrease in these strengths was less prominent. As temperature decreased, the yield and tensile strength values converged to a single value. For Heat A and B the strengths converged to a single value at approximately -250°F.

Figure 5.16 plots the yield strength for Heats A and B together. The yield strengths of both heats behaved the same with temperature, and have very similar values, until -250°F is reached. At, and below -250°F, the yield and tensile strength take on the same value for each heat.

Figure 5.17 plots the tensile strength for Heats A and B together. The tensile strengths of both heats also behaved the same with temperature, however Heat A was consistently stronger than Heat B at every test temperature. While Heat A is stronger, it is worth noticing, from Tables 5.5 and 5.6, that Heat B has

greater ductility at every test temperature as measured by percent elongation.

Both Heats A and B exceeded the minimum room temperature requirements specified by AWS for this particular weld filler material. Heat A filler material also exceeded the minimum room temperature requirements for the base material (ASTM A508 Class 4). Heat B exceeded all of the same requirements, except the tensile strength of Test B11 (103.3 Ksi) was slightly under the ASTM 105 Ksi minimum value. In general, the weld metal tensile properties matched the base material.

Figure 5.14

Heat A Strength Versus Temperature

Figure 5.15

Heat B Strength Versus Temperature

Figure 5.16

Heat A and B Yield Strength Versus Temperature

Figure 5.17

Heat A and B Tensile Strength Versus Temperature

Table 5.5
Heat A Tensile Properties

Specimen	Temp. (°F)	Yield (Ksi)	Ultimate (Ksi)	% RA	% Elong
1	650	84.7	101.4	66	27
2	650	87.0	103.1	72	24
3	0	101.7	113.7	70	26
4	-150	113.7	128.3	67	28
5	-165	120.8	134.2	66	28
6	-175	-	137.6	64	27
7	-200	121.0	137.3	47	27
8	-225	133.3	141.1	64	25
9	-250	148.5	148.5	64	25
10	-320	177.8	177.8	54	24
11	75	100.9	110.1	50	23
12	75	99.1	109.1	73	25
ASTM	75	85 MIN	105-130	45	18
AWS	75	78 MIN	90-110	-	17

Table 5.6**Heat B Tensile Properties**

Specimen	Temp. (°F)	Yield (Ksi)	Ultimate (Ksi)	% RA	% Elong
1	650	81.5	96.0	76	34
2	650	84.7	96.8	75	32
3	0	100.1	106.7	74	37
4	-150	115.0	121.9	66	34
5	-165	119.0	124.2	62	35
6	-175	122.2	125.6	66	34
7	-200	121.6	125.2	66	35
8	-225	130.8	131.7	61	33
9	-250	139.5	139.5	66	33
10	-300	159.7	159.7	63	32
11	75	95.4	103.3	74	35
12	75	98.7	106.5	73.5	31
ASTM	75	85 MIN	105-130	45	18
AWS	75	78 MIN	90-110	-	17

5.5 Metallographic and Chemical Analyses

Figures 5.18 and 5.19 display narrow groove welds made with Heat A and Heat B, respectively, using the final weld procedure. The weld fusion zone of each weld is refined. Notice in both figures that the top weld beads are not refined, since they have not seen thermal cycling from subsequent weld beads. Small regions in each weld have some coarse grains. However, when compared to Figure 5.1, which was a development weld, the microstructure revealed in Figures 5.18 and 5.19 is much more homogeneous.

From each of the above welds, samples were removed for chemical analysis. Figure 3.10 shows where the samples were located. Samples were removed in the weld fusion zone only to characterize as-deposited weld metal chemistry and dilution. An additional sample was taken in the base material for reference. Samples were removed from the bottom, middle, and top of the weld to determine through thickness variations. Table 5.7 lists the chemistry analysis results for a weld made with Heat A. Table 5.8 lists the same results for a weld made with Heat B. Also listed in the tables for comparison are the base material results and the weld wire chemistry for each heat.

In general, for both heats, the as-deposited chemistry did not vary much from the wire chemistry. The elements which varied the most from the wire chemistry are Mn, Cr and Ni and are due to the big difference between the wire and base material chemistries. Heat B contained greater concentrations of Ni

throughout the weld when compared to Heat A. Heat A, however, had greater concentrations of C, Cr, and Cu. The other alloying element concentrations were very similar. Note in particular that the tramp element concentrations (S and P) are still very low in the as-deposited weld metal.

In the through-thickness direction, both Heats A and B behave the same. This should be expected since the same weld procedure was used for welding all test mockups. Both heats appeared to have greater dilution with the base material in the bottom region based on the Ni and Cr concentrations. The bottom region is where the root and cold wire portion of the weld was made.

Figure 5.18

Heat A Narrow Groove Weld Metallography

Figure 5.19

Heat B Narrow Groove Weld Metallography

Table 5.7

Heat A As-Deposited Chemical Analysis

Element % Weight	Weld Wire	Top	Middle	Bottom	Base Material
C	0.14	0.12	0.15	0.13	0.21
Mn	1.60	1.49	1.45	1.44	0.25
Si	0.14	0.13	0.13	0.12	0.04
S	0.005	0.004	0.004	0.004	0.008
P	0.008	0.004	<0.002	0.003	0.008
Cr	0.10	0.21	0.24	0.26	1.64
Ni	0.90	1.00	1.00	1.04	3.28
Mo	0.53	0.51	0.51	0.51	0.50
Cu	0.19	0.15	0.12	0.12	0.08
Fe	Rem	Rem	Rem	Rem	Rem

Table 5.8

Heat B As-Deposited Chemical Analysis

Element % Weight	Weld Wire	Top	Middle	Bottom	Base Material
C	0.11	0.09	0.10	0.09	0.21
Mn	1.66	1.48	1.46	1.46	0.25
Si	0.16	0.16	0.15	0.15	0.04
S	0.005	0.005	0.005	0.006	0.008
P	0.005	0.005	0.004	0.003	0.008
Cr	0.05	0.08	0.09	0.12	1.64
Ni	0.97	1.04	1.09	1.12	3.28
Mo	0.54	0.51	0.51	0.51	0.50
Cu	0.02	0.02	0.02	0.03	0.08
Fe	Rem	Rem	Rem	Rem	Rem

6.0 Discussion and Conclusions

6.1 Discussion

This section thoroughly discusses the test results, tying together the effects that weld procedure, filler metal chemistry, and microstructure have on mechanical properties. Results from this project will also be compared to work in the literature. Particularly, fracture toughness, Charpy V-notch, and tensile test results will be compared to available ASTM A508 Class 4 base material and ASA weldments. References 3, 6 and 7 were used, since they contained excellent databases for ASTM A508 Class 4 and 1% Ni filler material. The data in the literature used similar test techniques to those in this project, thus they are directly comparable.

6.1.1 Weld Procedure and Effects on Metallurgy

Using the weld procedure shown in Table 5.1 produced a totally refined acicular microstructure in the weld fusion zone independent of the weld wire heat. This is clearly shown in Figures 5.18 and 5.19 for Heats A and B, respectively. The heat input for all welds was 34 KJ/in. maximum. This low heat input, as compared to typical ASA welds, created a faster cooling rate which promoted the formation of acicular ferrite in the welds (1,14,30).

Wire feed speed had the greatest effect on weld metal refinement. Using a lower wire feed speed for the same heat input produced a flatter thinner weld bead. Figure 6.1 graphically compares the difference between a weld bead produced by a high and low wire feed speed. A high wire feed speed produced a

thicker weld bead causing the coarse grained microstructure along the weld fusion line as shown in Figure 5.1. With a thinner weld bead the homogeneous microstructure as shown in Figure 5.2 was produced. The difference in the fusion zone microstructure is easily understood when the metallurgy of transformation-hardenable materials is considered. The metallurgical regions commonly associated with high strength low alloy (HSLA) weld heat affected zones (HAZ's) was previously discussed in section 1.2, and graphically presented in Figure 1.2. These regions are produced in any HSLA weld, independent of the weld process or joint design. Therefore, the metallurgy can be applied to narrow groove GTAW. Referring to Figure 1.2 the metallurgical regions of the HAZ are:

- a) The as-deposited columnar structure of the weld metal.
- b) The grain coarsened area adjacent to the fusion boundary associated with austenitic grain growth during the thermal cycle.
- c) The grain refined area where an austenitic microstructure has formed during the thermal cycle but grain growth has not been significant.
- d) The partial grain refined region where the thermal cycle was sufficient enough to refine most of the coarse grains.

In multipass welding the successive weld beads result in multiple overlap of these regions in the weld fusion zone. Overlap will also occur in conventional (non-narrow groove) welding. However, the important point to note about narrow groove multipass GTAW is that it provides better process control offering far more

accurate weld bead placement which allows control of the degree of refinement. Figures 6.2 and 6.3 shows the effect weld bead thickness has on weld fusion zone transformation. Each figure schematically shows four individual weld beads deposited in the narrow groove. Associated with each bead is the metallurgical regions discussed above and shown in Figure 1.2. Comparing the figures, it is shown that as weld bead thickness increased the proportion of untransformed columnar dendritic weld metal, and coarse-grained microstructure increased in the weld fusion zone. Therefore, it is highly desirable to weld with beads which are flat and as wide as possible to form a refined HAZ that extends over the entire layer thickness of the previous bead. The thin beads produced by using the weld procedure from Table 5.1 produced the homogeneous, refined microstructure shown in Figures 5.18 and 5.19, and was necessary to enhance mechanical properties.

Typical weldments are not homogeneous and contain a gamut of microstructures. Each microstructural region would be characterized by different mechanical properties. For example, comparing the Charpy V-notch testing of the refined and coarse grained regions from Table 5.4 the NDT_{35} ft.-lbs. was influenced by the microstructure. It is therefore imperative that when a weld does not contain a nearly homogeneous microstructure that all regions are characterized for mechanical properties so the "weakest link" can be found. These properties will have the largest influence any design. Producing a homogeneous microstructure

throughout the weld fusion zone caused good mechanical properties to be consistently obtained.

The weld procedure was characterized by a 3.2 lbs/hr maximum deposition rate. This is much lower than other weld processes shown in Table 1.2. However, the narrow groove weld joint does decrease the overall volume and would reduce the total fabrication time for completing the weld. Overwhelmingly, it is clear that the narrow groove joint design was selected more for bead placement process control than deposition rate. The synergic affects of the narrow groove joint design, low heat input and thin weld beads produced a refined, homogeneous microstructure characterized by high ductility, strength and toughness.

Figure 6.1**Comparison of Thick and Thin Weld Beads**

Figure 6.2**Thick Weld Beads and HAZ in Narrow Groove**

Figure 6.3**Thin Weld Beads and HAZ in Narrow Groove**

6.1.2 Fracture Toughness

The data from each heat had some scatter, but Heat B consistently had better fracture toughness than Heat A, especially in the transition regime. The microstructure of the welds made with Heat A and Heat B were the same as evidenced in Figures 5.18 and 5.19, respectively. Tables 5.7 and 5.8 show the as-deposited chemistry for welds made with each heat. Notice that the Heat B welds contained greater Ni and as discussed in section 4.2, Ni concentration directly affects notch toughness. Since the microstructures are similar, the higher Ni content seems the plausible reason for Heat B to having better fracture toughness.

Shown in Figure 5.3 is a conservative fracture toughness curve. Based on this curve the weld metal would be on the upper shelf and be ductile at -150°F . This weld material could, therefore, be used to very low temperatures without the risk of brittle fracture. Once on the upper shelf the material exhibits a toughness of 200 Ksi $\sqrt{\text{in}}$. and would have a great resistance to ductile tearing and large flaws.

For comparison, fracture toughness data of ASTM A508 Class 4 base material is shown in Table 6.1. This data was taken from Reference 6, which completely characterized different pressure vessel steels. The base material data is also plotted against the Heat A and B data, Tables 5.2 and 5.3, in Figure 6.4. Also drawn in the figure is the fracture toughness curve. The Heat A and B welds have better fracture toughness than the base material. Notice most of the base material data falls below the conservative fracture toughness curve, which was

based on the Heat A and B data. This is important since now the base material would be more limiting in fracture toughness than the welds, based on this comparison.

Figure 6.4

Fracture Toughness Comparison to Base Material

Table 6.1
Fracture Toughness Data from Reference 6

Temperature (°F)	K_{I0} (Ksi√in)
-200	78
-200	76
-200	120
-50	169
-50	151
-50	139
72	165
72	159
72	122
72	170
72	125
72	178
72	144
72	151
72	162
72	183
72	177
212	145
212	145
212	155

6.1.3 Charpy Impact

Charpy V-notch results from the development welds indicate that microstructure affects the NDT_{35} ft.-lbs. value. Referring to the results shown in Table 5.4 reveals that a refined microstructure had a 22°F improvement over the coarse-grained microstructure. As stated previously the NDT_{35} ft.-lbs. value is an indicator of the materials resistance to brittle fracture. The lower the NDT_{35} ft.-lbs. value the more resistant to brittle fracture the material will be. Therefore, the coarse-grained region of the development welds was less resistant to brittle fracture than the fine-grained region. Since a typical weld will have a wide range of microstructures each region will have a different resistance to brittle fracture. Therefore, any test which evaluates a materials resistance to fracture needs to sample a wide range of microstructures and assign a value to each region. This has the added value of determining which region is the most critical, the "weak link". With this information, the effects of welding variables can be evaluated with the ultimate objective of producing welds with the highest toughness. Thus, it was very desirable to have the fully refined acicular microstructure of the final developed welds, since the microstructure was homogeneous and very tough as evidenced by the final Charpy results.

Table 5.4 also compares the Heat A and B Charpy test results. A good comparison of how individual weld wire heats affect NDT_{35} ft.-lbs. can be made because three curves were generated for Heats A and B using the same weld

procedure. Based on the metallography the microstructures were the same and; therefore, the major difference between the heats would be the slight as-deposited chemistry difference shown in Tables 5.7 and 5.8. As shown in Table 5.4, Heat B consistently out performed Heat A in both NDT_{35} ft.-lbs. and USE. Referring to the as-deposited chemistries for each heat the reason why Heat B out-performed Heat A becomes clear. Heat B has higher Ni and lower C content in the weld metal. This combination is known to have a direct impact on material toughness (1). Therefore, other things being equal (i.e., weld procedure, microstructure) it is highly desirable to have a weld filler material which has a Ni content as high as possible, and a C content as low as possible to maximize toughness. Of course, the limits will be set by the specification of the wire procured. However, with this important information a materials engineer can handpick a weld wire heat from the available heats which has this optimal combination.

Table 6.2 lists available Charpy V-notch data from the literature and the average values from Heats A and B. NDT_{35} ft.-lbs. and USE are listed for each material. The data from the literature was taken from References 3 and 7 because both references investigated the affects irradiation has on Charpy impact toughness of ASTM A508 Class 4 base material and 1% Ni ASA welds. Therefore, the data is very useful in establishing if the GTAW narrow groove procedure is capable of performing as well as these materials. The NDT_{35} ft.-lbs. and USE values are also graphically presented in the bar charts of Figures 6.5 and 6.6,

respectively. The materials are grouped in colors for the readers ease. Notice first from Figure 6.5 that the automatic submerged arc welds (ASAW) were never as good as the base material in terms of NDT_{35} ft.-lbs. and; therefore, would be considered the "weak link". The GTAW narrow groove process using Heats A and B outperforms the ASAW process. Additionally, the GTAW narrow groove NDT_{35} ft.-lbs. is equivalent to the best ASTM A508 Class 4 base material value. In fact, Heat B exceeded the best ASTM A508 Class 4 base material by 52°F.

Figure 6.6 shows a comparison of the USE. Except for A508 ID2, all of the remaining values are very similar. Notice how the Heats A and B narrow groove GTAW values are particularly close to the ASAW values. This would indicate that the USE is not dependent on microstructure. Clearly the NDT_{35} ft.-lbs. and USE data indicates that a GTAW procedure can be used which will outperform ASAW and at least meet the base material properties. Again, this re-enforces that a GTA narrow groove weld using this procedure will not be the "weak link" for these material combinations based on Charpy impact properties.

Figure 6.5

NDT₃₅ ft.-lbs. Comparison

Figure 6.6

USE Comparison

Table 6.2

Charpy V-notch Test Data Comparison

Material	NDT ₃₅ (°F)	USE (ft.-lbs.)
Heat A average	-133	137.7
Heat B average	-184	160.6
A508 124S285 ³	-132	152
A508 122Z195 ³	-87	138
A508 ID1 ⁷	-115	157
A508 ID2 ⁷	-110	101
ASAW 289246 ³	-65	137.9
ASAW 624116 ³	-63	137.9
ASAW IP2915 ⁷	-50	140

6.1.4 Tensile Tests

Tensile test results showed that Heat A had a slightly higher tensile strength than Heat B at every temperature tested, as shown on Figure 5.17. While Heat A was stronger, Heat B was better in ductility as indicated by the % elongation listed in Tables 5.5 and 5.6. Again, since the microstructures were the same, the difference in strength and ductility was due to the as-deposited chemistry difference. Tables 5.7 and 5.8 showed that Heat A had a higher C and Cr content in the weld metal. These elements enhance material strength, but adversely affect ductility as was discussed in section 4.2.

Both heats exceeded the minimum AWS requirements for strength and ductility and; therefore, could be used in conjunction with the GTAW narrow groove process.

Table 6.3 lists available tensile data from the literature and room temperature data from the base material, Table 4.2, used in this project (heat 91H-005). The data taken from the literature is from Reference 6.

Figure 6.7 plots the yield strength data from the literature and base material heat 91H-005 with Heats A and B. In general, the yield strengths of all the materials were similar. The base material data is grouped within the weld material with no noticeable difference between them.

Figure 6.8 plots the tensile strength data from the literature and base material heat 91H-005 with Heats A and B. When compared to the weld values the

base material has slightly higher values, especially when compared to Heat B. This is to be expected since Heat B had lower C and Cr content and; therefore, would have lower strength.

Based on the Heat A and B tensile results and comparison to base material it would not be advisable to use the base material tensile strength when performing a design analysis of a welded structure. This would give an unconservative estimation of the performance of the weld. Instead, the yield strength of the weld should be used for calculations because these values are more tightly grouped with the base material.

Figure 6.7

Yield Strength Versus Temperature Comparison

Figure 6.8

Tensile Strength Versus Temperature Comparison

Table 6.3**Tensile Data from Reference 6 and Heat 91H-005**

Material	Temp (°F)	Yield (Ksi)	Ultimate (Ksi)	% RA	% Elong
Ref 6	-200	116.6	138.1	-	63.2
Ref 6	-200	117.3	137.4	-	63.1
Ref 6	-130	100.0	130.8	24.8	66
Ref 6	-130	107.5	130.3	24.8	65.6
Ref 6	-50	104.6	122.6	22.8	68.1
Ref 6	-50	103.0	121.1	21.8	63.4
Ref 6	72	95.5	110.5	22.7	68.0
Ref 6	72	94.0	109.2	24.0	68.9
Ref 6	212	94.4	108.8	21.2	69.3
Ref 6	212	91.9	105.3	20.0	71.5
91H-005-1	75	99.5	117.4	25.0	71.9
91H-005-2	75	99.5	115.9	25.0	74.2
91H-005-3	75	98.6	115.0	26.0	74.9

6.1.5 Metallographic and Chemical Analyses

Figures 5.18 and 5.19 display narrow groove welds made with Heats A and B, respectively. The entire weld fusion zone is refined and completely homogeneous. No lack of fusion or weld defects were found in any of the metallographic samples. Thus indicating that the weld procedure has good process control and margin for preventing these types of defects. Notice that the top beads in both figures were coarse because they were not subjected to any additional thermal cycles, as discussed in section 6.1.1, to refine this structure.

Figure 6.9 shows photomicrographs of a typical weld made with Heat A near the weld fusion line, Heat B microstructure was the same. The microstructure is characterized by very small grains of acicular ferrite with a small amount of grain boundary ferrite. The weld procedure met the objective of producing a fully refined fusion zone with an acicular ferrite microstructure. This is the dominant reason the weld mechanical properties were so good. For comparison, Figures 6.10 and 6.11 show photomicrographs near the weld fusion line from the procedure development weld shown in Figure 5.1. Notice how big the coarse grains are when compared to the fine grains of Figure 6.9. This coarse microstructure is characterized by ferrite with an aligned second phase of carbides.

All microstructure characterizations were done using references 34 and 35. Obviously, this coarse grained microstructure was not optimal for maximizing material properties as was found by the Charpy V-notch test results for the

development welds. The final weld procedure was capable of producing a totally refined acicular microstructure by applying wide thin weld beads to cause thermal refinement in the entire fusion zone.

Tables 5.7 and 5.8 list the as-deposited chemistry through the thickness for welds made with Heats A and B. Both Heats A and B exhibited minimal dilution with the base material. Notice that the tramp element concentrations are low in the as-deposited weld metal due to the initially low values in both wire heats, and the base material (91H-005). The elements which showed the most dilution were Mn, Cr and Ni. This is due to the large difference in the base material and wire chemistry for these elements. Even with these large differences the dilution is very low, which implies that the weld procedure has good control over as-deposited chemistry due to the low heat input. The low heat input caused minimal base material melting into the weld puddle and; therefore, the base material had only a small influence on the final as-deposited chemistry. This is significant and important especially if the weld procedure was used on a material which had higher tramp element (S, P) concentrations. These elements would influence the welds performance by potentially causing solidification cracking if dilution was high enough.

Table 6.4 provides a good representation of how composition affects mechanical properties. Listed in the table are the major elements which affect mechanical properties based on the data in Tables 5.7 and 5.8. Also listed are the

average tensile and Charpy V-notch results for each heat. Since the weld procedure was a constant, the microstructure was the same independent of the weld wire heat. This was confirmed by metallography as discussed above. Heat A contained higher levels of C and Cr, but lower Ni than Heat B. Comparing the mechanical properties, Heat B was better in toughness and ductility, but lower in strength than Heat A. Ni is a known enhancer of toughness, while C and Cr are known to increase strength. These effects were previously discussed in section 4.2. These results indicate that the allowable chemistry variations for a given weld filler specification can play a significant role in determining the final mechanical properties of a weldment. For example, with the knowledge that this procedure is capable of producing a fully refined microstructure, a materials engineer can preselect a certain heat of wire to either perform for strength, toughness, or a modest combination of both based on wire chemistry.

Figure 6.9

Typical Weld Microstructure from a Final Weld

100x magnification

425x magnification

Figure 6.10

**Typical Microstructure from a Procedure
Development Weld**

50x magnification

Figure 6.11

**Typical Microstructure from a Procedure
Development Weld at High Magnification**

210x magnification

425x magnification

Table 6.4**Chemistry and Mechanical Properties Comparison**

Heat	A	B
Average C% (as-deposited)	0.13	0.09
Average Cr% (as-deposited)	0.24	0.10
Average Ni% (as-deposited)	1.01	1.08
Tensile at RT (Ksi)	109.6	104.9
Ductility at RT (% elong)	24	33
NDT ₃₅ ft.-lbs. (°F)	-133	-184
USE (ft.-lbs.)	137.7	160.6
Weld Microstructure	Acicular Ferrite	Acicular Ferrite

6.2 Conclusions

1. The objectives of this thesis were met. A narrow groove GTA welding procedure was developed which fully refined the weld fusion zone to produce an acicular ferrite microstructure. Knowledge of HSLA steel metallurgy was successfully used in conjunction with understanding of the characteristics of the narrow groove process to produce mechanical properties better than ASAW and, as good, as the base material when measured by Charpy V-notch, tensile, and fracture toughness testing techniques.
2. All things being equal (i.e., weld procedure, microstructure) slight variations in weld wire chemistry within the acceptable specification limits have an impact on mechanical properties. Heat B clearly outperformed Heat A measured by ductility and toughness because it had greater Ni concentration. Heat A had greater tensile strength because of its higher C and Cr levels.
3. The initial reason for selecting GTA narrow groove welding was to improve the weld joint completion rate. However, it has become very clear that narrow groove also offers tighter control over how and where weld beads are deposited. This improved process control allows metallurgical control of

the weld joint properties by improving total microstructure refinement within the weld fusion zone. In the long-term, this seems to be a far more valuable development than improved joint completion rates.

4. Thin flat weld beads are preferred because of the more extensive weld heat affected zone overlap which causes increased refinement in the weld fusion zone.

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