The Development of a Ferritic Consumable for Welding Grain-Refined Fe-12Ni-0.25Ti to Retain Toughness at 4.2 K

A matching 14% ferritic filler metal is used with the gas tungsten arc process

By H. J. Kim and J. W. Morris, Jr.

ABSTRACT. Multipass gas tungsten arc welding (GTAW) has been adapted for welding grain-refined Fe-12Ni-0.25Ti steel with a matching ferritic 14% filler metal. The weld is made as a multipass weldment using a single V-groove preparation at heat inputs from 7-17 kJ/cm (18-43 kJ/in.).

The ferritic weldment has a strength roughly matching that of the base metal and exhibits excellent toughness both in the weld metal and in the heat-affected zone at temperatures as low as liquid helium temperature (4.2 K, i.e., -452°F).

The excellent toughness is attributed to two factors: the chemical cleanliness of the GTAW deposit, and the refined microstructure of the weldment and the retention of fine microstructure in the heat-affected zone. Grain refinement is accomplished by the sequential rapid thermal cycles experienced by the material during multipass GTAW welding.

Introduction

Previous research (Ref. 1-3) in this laboratory has demonstrated that the ductile-brittle transition temperature of Fe-(9-12%)Ni ferritic cryogenic steels can be suppressed to below 4.2 K (-268.8°C or -452°F) by applying heat treatments which establish a rather precise control over the alloy microstructure. Subsequent welding introduces the simultaneous problems of establishing a suitable microstructure in the weld and retaining an appropriate microstructure in the heat-affected zone. Initial research on the welding of this (Ref. 4) and similar alloys (Ref. 5) for 4 K (-452°F) service employed high-nickel Inconel filler metals to avoid the problem of brittleness in the weld deposit. However, the GMA welding of Fe-12Ni-0.25Ti with Inconel 92 (i.e., AWS ERNiCrFe-6) filler metal (Ref. 4) was only partly successful. The ERNiCrFe-6 filler metal was inferior to the ferritic base metal over the whole temperature range, not only in its yield strength but also in its notch toughness. In addition, the ERNiCrFe-6 weldment exhibited a fusion zone brittleness which is familiar from other research (Ref. 6) on the welding of Fe-9%Ni steel with non-matching austenitic filler metals; cracks starting in the heat-affected zone tend to propagate along the fusion line.

Recently, researchers at Nippon Kokan K.K. and Kobe Steel (Ref. 7) jointly developed an automatic GTAW variation which uses a matching ferritic filler metal to weld quenched-and-tempered 9Ni steel for service at 77 K (-321°F). The success of this procedure is largely due to two advantages of multipass GTAW welding:

1. The low level of air contamination in the weld deposit.
2. The cyclic heat treatment of the weld deposit by subsequent passes.

Cyclic heat treatment is at least partly

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controllable through independent variation of the heat input and filler metal feed rate. It is, as a result, possible to obtain a terrific weld deposit which is both clean and relatively fine-grained and which, consequently, has good toughness at 27 K (−321 °F). Preliminary work at the NASA Lewis Research Laboratory (Ref. 8) suggests that Fe-12Ni-0.25Ti may also be GTA welded with matching filler metal for 77 K (−321 °F) service.

The microstructural constraints which must be satisfied to achieve good toughness at 4 K (−452 °F) are much more stringent than those required for toughness at 77 K (−321 °F). Nonetheless, the metallurgical approach of the multipass GTAW process as a means of controlling the microstructure of weldments together with the success already achieved at 77 K (−321 °F) suggest that this process may be successful for joining ferritic cryogenic steels for 4 K (−452 °F) service.

Some very promising initial results of research on the multipass GTA welding of Fe-12Ni-0.25Ti steel with matching ferritic filler metal are described below.

### Alloy Preparation and Experimental Procedure

#### Alloy Preparation

The compositions of the base metal and filler metals used in this research are given in Table 1. The base metal has a nominal composition, in weight percent (wt-%), of Fe-12Ni-0.25Ti. The two filler metals (“A” and “B”) have a slightly higher nickel content (14 wt-%) plus intentional additions of other alloying elements. The increase in nickel content was made in the hope of lowering the ductile-to-brittle transition temperature of the weld metal (Ref. 9, 10). Boron was added to filler metal B in the expectation that it would act as a grain boundary surfactant to promote grain refinement and inhibit intergranular or intercellular fracture (Ref. 11, 12). The Mn and Si additions to filler metal A were made to explore their effect on mechanical properties since these are common additions in commercial heats.

Both the base and filler metal alloys were cast, after vacuum-induction melting, from pure starting materials. The base metal was cast as a 25 lb (9.425 kg) ingot, homogenized at 1200 °C (2192 °F) for 24 hours (h) under inert atmosphere, and then upset cross-forged at 1100 °C (2012 °F) into plates, each 7 in. wide and 0.6 in. thick (178 × 15 mm). Each plate was then cut in half and heat-treated through the “2B” thermal-cycling treatment diagrammed in Fig. 1. This treatment, which has been described elsewhere (Ref. 1), establishes a fine microstructure of nearly equiaxed grains of ~1 μm mean diameter. The plates were then machined into the joint configurations described below.

The weld filler metal was cast into 5 lb (9.25 kg) 1 in. (25 mm) diameter ingots. After homogenization at 1200 °C (2192 °F) for 24 h, the ingots were hot rolled into ½ in. (7.9 mm) square bars at a starting temperature of 1100 °C (2012 °F), then hot swaged to ½ in. (6.4 mm) diameter round bars and surface ground to remove any oxides. The ground bars were cold swaged to 3/4 or 7/8 in. (1.95 or 2.38 mm) diameter bars, which were used as the filler metal for manual GTA welding.

#### Thermal Cycling Studies

Thermal cycling studies were conducted to gain insight into the response of the alloy to the rapid heating and cooling cycles encountered during welding. All thermal cycling studies were done on the base alloy, Fe-12Ni-0.25Ti, in one of the two starting conditions:

1. The as-annealed condition, which was used as a crude representation of the initial condition of the weld deposit.
2. The “2B” condition, which represented the initial state of the heat-affected zone.

The thermal cycles were imposed on slightly oversized Charpy impact specimens in an induction furnace. The sample was suspended in the center of a 38 mm (1 ½ in.) diameter quartz tube surrounded by the induction coil. For cycles having a peak temperature lower than 1200 °C (2192 °F), the temperature was monitored by a chromel-alumel thermocouple connected both to a chart recorder and to a digital indicator. For cycles having peak temperature higher than 1200 °C (2192 °F), an optical pyrometer was used.

Once the selected peak temperature had been reached, the sample was dropped into a quenching bath of either water or oil. An example temperature-time profile for a 1300 °C (2372 °F) peak temperature cycle with an oil quench is sketched in Fig. 2.

#### Welding Procedure

The plates to be welded were machined into one of the two configurations diagrammed in Fig. 3—a 45 deg single bevel or a 60 deg single-V-groove. The plates were welded with a multipass GTAW procedure using one of the two sets of welding conditions tabulated in

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**Table 1—Compositions of Base Metal, Filler Metals and Weld Metals, %**

<table>
<thead>
<tr>
<th></th>
<th>Fe</th>
<th>Ni</th>
<th>Ti</th>
<th>Mn</th>
<th>Si</th>
<th>P</th>
<th>S</th>
<th>O</th>
<th>N</th>
<th>B</th>
<th>C</th>
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<tr>
<td>Base Metal</td>
<td>bal.</td>
<td></td>
<td></td>
<td>12.07</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td>0.15</td>
<td></td>
<td></td>
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<tr>
<td>Filler A</td>
<td></td>
<td>14.04</td>
<td>0.21</td>
<td></td>
<td></td>
<td>0.37</td>
<td>0.09</td>
<td>0.005</td>
<td>0.002</td>
<td></td>
<td></td>
</tr>
<tr>
<td>Filler B</td>
<td></td>
<td>13.87</td>
<td>0.16</td>
<td></td>
<td></td>
<td>0.33</td>
<td>0.08</td>
<td>0.005</td>
<td>(a)</td>
<td></td>
<td></td>
</tr>
<tr>
<td>Weld A</td>
<td></td>
<td>13.98</td>
<td>0.20</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
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<tr>
<td>Weld B</td>
<td></td>
<td>13.81</td>
<td>0.15</td>
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<td></td>
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<td></td>
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</tr>
</tbody>
</table>

(a) Negligible, (b) Not analyzed.

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**Table 2—Welding Conditions**

<table>
<thead>
<tr>
<th>Heated input, kJ/cm</th>
<th>Arc voltage, V</th>
<th>Welding current, A</th>
<th>Welding speed, cm/s</th>
<th>Shielding gas flow rate, clfh</th>
<th>Root opening</th>
<th>Interpass temperature, °C</th>
</tr>
</thead>
<tbody>
<tr>
<td>1500-1800</td>
<td>15-18</td>
<td>150-1800</td>
<td>0.4</td>
<td>Pure argon, 25</td>
<td>Same as welding rod diameter</td>
<td>50-150</td>
</tr>
<tr>
<td>17-18</td>
<td>18-20</td>
<td>250-300</td>
<td>0.4</td>
<td></td>
<td></td>
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</tbody>
</table>
Weld Characterization

The completed single-V-groove weld in a 0.6 in. (15 mm) plate required ~35 passes at a heat input of 7 kJ/cm (18 kj/in.) vs. ~12 passes at a heat input of 17 kJ/cm (43 kj/in.).

Bead-on-plate welds were also made under a variety of conditions to study deposition characteristics and ferritic weld structures.

Table 2. Completing the single-V-groove weld in a 0.6 in. (15 mm) plate required ~35 passes at a heat input of 7 kJ/cm (18 kj/in.) vs. ~12 passes at a heat input of 17 kJ/cm (43 kj/in.).

Mechanical Tests

The mechanical tests conducted included tensile, Charpy impact and fracture toughness tests. To prepare test specimens, the welded plates were sliced perpendicular to the joint and etched with 10% nital or 2% HF solutions to reveal the weld location on the sliced surface. The specimens were then examined nondestructively by x-ray radiography using a voltage of 250 kV and a 10 mA current. No significant discontinuities were found in these weldments.

The microstructures of the welded joints were studied by optical metallography and metallographic microscopy equipped with energy-dispersive x-ray analysis (EDAX). The scanning electron microscope was also used to examine nondestructively by x-ray radiography using a voltage of 250 kV and a 10 mA current. No significant discontinuities were found in these weldments.

In order to minimize the deviation of the fatigue pre-crack from the desired location, the Charpy specimens were given an 8 mil (0.008 in.) saw cut of about 1 mm (0.04 in.) depth at the root of the Charpy V-notch. After fatigue pre-cracking to initial crack length-to-specimen dimensions, the specimens were broken at 77 K (—321 °F) under immersion in liquid nitrogen at a testing temperature of 0.06 mm/min (0.002 ipm).

Two compact tension fracture toughness specimens of the weld metal were also tested at 4-5 K (~—450 to —450°F), again using a styrofoam box modification of the lucite box technique reported previously (Ref. 2). Space for the compact tension specimen was hallowed out of a 25 mm (1 in.) thick styrofoam block, and grooves were made on the inside surfaces to facilitate helium flow. The
specimen assembly was wrapped and inlet holes were provided for liquid helium as described in the literature (Ref. 2). The temperature during the sample cooling and testing was monitored by a Au + 0.07Fe-Chromel thermocouple embedded in the sample near the crack tip. The fracture toughness test was conducted after the sample temperature had stabilized near 4 K (−452°F).

None of the fracture toughness specimens tested met ASTM thickness requirements for plain strain conditions. The plain strain toughness values were hence estimated from the “equivalent energy” criterion (Ref. 16). The relevant equation is:

$$K_{IC} = \frac{P q}{f(a/w) \sqrt{A_1 A_2 \left(B W^3\right)^{-1}}}$$

where $a$, $w$, $B$ and $S$ are respectively the crack length, specimen width, thickness and span length, $P$ is the maximum load on the linear portion of the load-displacement curve, $f(a/w)$ is the geometric shape factor (Ref. 15) and $A_1$ and $A_2$ are the areas under the load-displacement curve up to the maximum load ($A_1$) and the load $P_q$($A_2$). The fracture toughness values given in this paper are the averages of two or more test results.

**Experimental Results**

*Weld Microstructures*

**Bead-on Plate Welds.** A variety of test weldments were made at heat inputs ranging from 7 to 17 kJ/cm (18 to 43 kJ/in.). These had essentially similar microstructures, which consisted of columnar grains made up of bundles of narrow solidification cells growing in the same direction. The columnar grains are revealed by chloride etching, as in Fig. 5B. They appear to grow epitaxially from half-melted coarsened grains along the fusion line. The average lateral dimension of those grains tends to increase near the center of the bead. The cellular solidification substructure of these grains is revealed by a nital etch, as in Fig. 5A. Each columnar grain contains a bundle of cells. No dendritic substructure was observed.

An examination of partially overlapped bead-on-plate weld passes showed that the solidification structure is hardly affected by the subsequent passes but the columnar grains are destroyed. The result is a fine equiaxed grain structure even at the fusion boundaries. The columnar grains or coarsened HAZ grains formed during the earlier pass did not coarsen further at the fusion boundaries of a latter pass.

**Full thickness weldments.** Figure 6 shows the macro- and microstructures of a completed weldment with a 7 kJ/cm (18 kJ/in.) heat input. The HAZ of each bead deposited reheats, recrystallizes and breaks up the original cast columnar microstructure—Fig. 6. Thus, the whole weldment is repeatedly transformed, refined and possibly tempered during fabrication. While a few isolated islands of partially-refined structure remain (Fig. 6), throughout the greater part of the weld volume both the coarsened HAZ and the large columnar grains are broken up to produce a fine-grained structure.

A similar grain refinement was attained with 17 kJ/cm (43 kJ/in.) heat input. The grain size of the irregular grains is in the range 5-10 μm in the well-refined regions.

**Heat Simulation Results**

Figure 7 shows the microstructural changes, and Fig. 8 shows the change in impact energy when as-annealed specimens are given rapid thermal cycles. The original as-annealed structure (Fig. 7A) has about 60μ grain size and fractures in a cleavage mode with an impact energy of ~10 ft-lb (14 J) at 77 K (−321°F). The specimens reheated to below the $A_s$ temperature (678°C i.e., 1252°F) retained the original grain size, low toughness and brittle fracture mode.

Cycling to near 800°C (1472°F) peak temperature destroys the original structure and creates non-uniform irregular grains—Fig. 7C. The grain size varies from 10 to 50μ. The impact toughness is dramatically improved, i.e., to ~150 ft-lb or 203 J (Fig. 8), and a dimple-rupture type fracture mode is established. When the cycle reaches a peak temperature >1200°C (2192°F), however, grain growth takes place (Fig. 7D) and leads to low toughness in a brittle mode (15 ft-lb or 20.3 J). It is interesting to note the wide range of peak temperatures over which.

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**Fig. 5 — Microstructures of bead-on-plate welds etched with: A —5% nital; B —acidified FeCl₃ solution**

**Fig. 6 — Macro- and microstructures of full thickness welded joint with 7 kJ/cm (18 kJ/in.) heat input.**
Mechanical Properties of the Weldments

The measured mechanical properties of the ferritic weldments are shown in Table 3 along with those of the base metal (Ref. 1).

**Tensile Properties.** The results of tensile tests of joints welded with filler metal B are shown in Table 3. The tensile properties in the transverse direction compared closely with those of the base metal in the case of the 7 kJ/cm (18 kJ/in.) heat input but were slightly lower for the higher heat input, 17 kJ/cm (43 kJ/in.). It was confirmed by etching the broken specimen that the fracture occurred outside the weld metal, perhaps near the base metal/HAZ boundary. Since the specimen was severely deformed in the fracture region, it was not possible to define an exact fracture site.

The tensile data indicate that the yield strength of the weld metal is the same or higher than that of the base metal. Room temperature microhardness tests also give higher values in the weld metal (Hv = 290 Kg/mm²) than in the HAZ (Hv = 280) or base metal (Hv = 260).

**Charpy V Notch Impact Toughness.** The variation of impact toughness with notch location at 77 and 4.2 K (-321 and -452°F) is shown in Fig. 11. The HAZ has the highest impact values and is about 50 ft-lb (68 J) higher than the base metal at both 77 and 4.2 K (-321 and -452°F), as observed in the HAZ simulation. The weld metal also has higher impact energy than the base metal. It fractures in a ductile mode at 77 K (-321°F) but sustains some quasi-cleavage at 4.2 K (-452°F).

The slight brittleness at 4.2 K (-452°F) in the single bevel-groove weld was completely suppressed in the single-V-groove weld. As shown in Fig. 4, filler metal B weld metal deposited in the V-groove with either of the two levels of heat input maintained high impact values at liquid helium temperature. SEM fractographs taken from broken specimens at 77 and 4.2 K (-321 and -452°F) show a completely ductile mode — Fig. 12. The impact DBTT (ductile brittle transition temperature) of weld metal obtained with filler metal B was, hence, successfully suppressed to below liquid helium temperature.

As shown in Table 4, the impact toughness of the filler metal A weld was much lower than that of filler metal B weld, particularly at 4.2 K (-452°F). The lower toughness of the filler metal A weld is, apparently, a microstructural effect. Examination of low toughness filler metal A weld specimen revealed two types of inclusion particles:

1. A small spherical type often found inside fracture dimples in the ductile region.
2. A large, round or irregular type which was often observed in the brittle region.

These two kinds of particles are shown in Fig. 13, with the EDAX (electron diffraction) analysis of each. The spherical particles within dimples always registered Ti and S while the irregular particles in the brittle region showed Mn, Si and Ti without any trace of S.
Table 4 also shows an effect of weld geometry: the DBTT of the filler metal A weld is lower than 77 K (−321 °F) in the single-V-groove weld but higher in the single-bevel-groove weld. This behavior apparently reflects the influence of weld geometry on grain refinement; the single-bevel-groove weld has a lower potential for grain refinement along the fusion boundary during multipass welding. Optical examination of the single-bevel-groove weld showed that the beads just beside the straight joint side retained the as-cast columnar structure, particularly near the root bead.

Fracture Toughness. Since the filler metal A weld tended to become brittle at 4.2 K (−452 °F) in Charpy impact tests, further fracture toughness tests were done only on joints welded with filler metal B. Figure 14 shows typical load-crack opening displacement (COD) curves obtained from three-point bend tests at 77 K (−321 °F) with three different locations of the fatigue crack. All the specimens seemed immune to unstable crack propagation; the specimens were fully plastic and the pre-induced cracks grew slowly in a stable manner until the test stopped. A value of $K_{eq} \approx 75$ ksi $\sqrt{in}$ was computed from these curves. The fracture toughness values computed by the “equivalent energy” method are presented in Table 5.

The results of the compact tension fracture toughness tests weld metal at 77 K (−321 °F) and 4–6 K (−452 to −450 °F)
The weld metal chemistry explored in this work differs from that of the base metal in three respects: the increased Ni content, the addition of Mn and Si (filler metal A) and the addition of boron (filler metal B).

**Nickel Content.** The nickel content was increased to 14 wt-% to err on the side of safety in achieving a low DBTT in the weld metal. It is not clear that this increase is necessary. While a satisfactory DBTT was obtained, the thermal cycle simulation studies suggest that satisfactory properties might also have resulted from the base metal composition.

Current metallurgical understanding does not permit an a priori choice of an optimum nickel content for low temperature toughness. While an increase in nickel content is often found to lead to a decrease in the DBTT (Ref. 9, 10), the mechanism of the effect is uncertain. However, it is likely that the primary benefit achieved from nickel is indirect, through the influence of nickel on the response of the microstructure to thermal treatment.

Previous research suggests that there is an optimum nickel content of the weld filler metal which is somewhat above that of the base metal. In the GTA welding of 9% steel an increase in nickel content from 5 to 11 wt-% gave a monotonic improvement in toughness at 77 K, i.e.,
-321°F (Ref. 17), but toughness deteriorated when the nickel content was increased further to 17 wt-%. The problem at higher nickel content may be associated with an unstable austenite retention in the weldment, as is apparently the case in 18%Ni (250 grade) maraging steel (Ref. 18).

Manganese and Silicon. Manganese and silicon are common deoxidizers which were added to filler metal "A" to determine their effect on the final weldment properties. Their presence led to a deterioration in the weld metal toughness, and this seems to be clearly associated with the formation of large (Mn-Si) oxides.

These results are in general agreement with previous research. Silicon is often intentionally added to ferritic weldments for 9Ni steel (Ref. 19) since 9Ni contains a significant alloying addition of manganese, and it is known that a low Mn/Si ratio is needed to obtain efficient removal of deoxidation products to the weld surface (Ref. 20). Moreover, serious porosity problems have been encountered in ferritic GMA welding of 9%Ni steel with Si-free weld metal (Ref. 21) and in the welding of low carbon manganese steels (Ref. 20), apparently because of insufficient deoxidation in the weld deposit. It has, however, been found that
both Si and Mn decrease the low temperature toughness of welded 9Ni steel (Ref. 17). With Si-containing weld chemistries, large (Mn,Si) oxides have been found on the fracture surfaces of ferritic weldments in 9Ni steel (Ref. 22), and apparently cause embrittlement by mechanisms similar to that noted in the present work.

The results obtained with filler metal “B” in the present work show that manganese and silicon are not necessary alloy additions to the weld metal. The 12Ni base metal contains no manganese; deoxidation and scavenging are, apparently, efficiently accomplished by the alloy addition of Ti, as inferred by previous researchers (Ref. 1, 10). As shown in the auger analysis of inclusion particles presented as Fig. 17, Ti acts as a deoxidizer and cleanser of carbon and sulfur to getter other potentially deleterious elements. Given the relative cleanliness of the GTA welding process, the titanium content of the base metal seems sufficient to ensure a clean and well-gettered weldment.

**Boron.** Boron was added to filler metal “B” in the expectation that it would act as a strong surfactant in boundaries of the weld metal to improve grain refinement and prevent intergranular separation, as it does in other alloys (Ref. 11, 12). Its actual role in the weldment has not been identified since no trace of boron was detected by auger analysis of the microstructure of fracture surface. Boron did, however, seem to be beneficial to the welding process. The addition of a small amount of boron to the Fe-12Ni-0.25Ti base plate improved weld metal fluidity in bead-on-plate welding. Also it was found that filler metal “B”, which contained 30 ppm boron, produced a more easily controlled weld puddle and better weld beads than did the boron-free filler metal “A”.

**Effect of Weld Thermal Cycles**

The high impact toughness and the low ductile-brittle transition temperature of the weldment is a consequence of the multpass GTA welding process, which appears to refine the grain size of the weld deposit and to retain fine grain size in the heat-affected zone. A detailed transmission electron microscopic analysis of the consequences of the multpass GTA process is now in progress. Some important aspects of the grain refinement are, however, clear from the evidence obtained in the work reported here.

In the conventional treatment of ferritic Fe-Ni cryogenic steels there are two mechanistically different ways of establishing a fine effective grain size; these are:

1. The direct crystallographic refinement of the martensite or ferrite structure.
2. The introduction of a fine distribution of precipitated austenite which serves to break up the crystallographic alignment of the martensite or ferrite structure.

In the present case it appears that the grain refinement mechanism is direct crystallographic refinement of the deposit, since no measurable retained austenite is detected in the weldment or heat-affected zone, even at room temperature. This result is in agreement with previous research on the GTA welding of 9Ni steel with ferritic filler metals (Ref. 21), but is at variance with the results of similar research by Tamura, et al. (Ref. 23). Here, Tamura, et al. reported a significant amount of thermally stable austenite (approximately 6%) in the weld metal, and suggested a correspondence between this austenite and the improvement in impact toughness. The improvement in toughness obtained in the present work is associated with microstructural changes in the ferritic deposit during the fast heating and cooling cycles associated with multipass welding.

There is a reasonable body of prior research showing that a rapid austenitizing cycle can accomplish a significant grain refinement of steel and improve its mechanical properties with respect to conventionally treated material. This research involved iron-carbon alloys (Ref. 24, 25) and HY-130 steel (SNi-Cr-Mo-B) (Ref. 26).

Research by Porter, et al. (Ref. 26) on HY-130 steel is particularly relevant, since these workers found that the degree of grain refinement increased with heating rate and also noted changes in dislocation substructure which were more pronounced in specimens given a rapid thermal cycle. The reheating rate in multipass GTA welding is quite rapid and was shown above to lead to an apparent grain size in the 5-10 μm range. The substructural changes caused by the rapid thermal cycle must, however, be important since the 5-10 μm apparent grain size of the weldment is considerably above that (less than 1 μm) which appears to be necessary to lower the ductile-brittle transition of the base metal below 4.2 K (-452°F) in conventional heat treatments.

**Alternative Welding Processes**

The GTA welding process was selected for the present research because of its cleanliness and its controllability. In addition, the large, uniform heat-affected zone of the GTA process is useful to ensure the cyclic heat treatment of previously deposited material. This advantageous combination is not easily achieved in other welding processes such as gas metal arc (GMA) or electron beam (EB) welding. Conventional electron beam welding has already been demonstrated (Ref. 4) to be inappropriate because of the coarse solidification structure established in the weld metal. The GMA process also presents difficulties but deserves further exploration because of the economic benefits to be achieved from its higher deposition rate.

The gas metal arc welding process suffers from two shortcomings:

1. Localization of its heat-affected zone due to its high deposition rate and its bell-like bead shape.
2. Its relatively high contamination, particularly by oxygen and nitrogen, which arises from plasma jet instabilities in the consumable electrode process.

Zanis, et al (Ref. 27) considered the problem of achieving grain refinement in a high-rate deposition GMAW process and have proposed reheatment treatments using heating sources such as autogenous GTAW or lasers to refine the GMA weld deposit. Dolby (Ref. 28) has also proposed in-process techniques to refine the microstructure of welds. But the contamination problem must also be overcome. Watanabe, et al. (Ref. 29) investigated the notch toughness of ferritic SMAW weld metal and found that the toughness decreased dramatically when the oxygen content rose above 100 ppm. Further research is necessary before the GMAW can be successfully used for welding ferritic steels with ferritic consumables for deep cryogenic applications.

**Conclusions**

1. The GTAW process can be successfully used to weld grain-refined Fe-12Ni-0.25Ti alloys with ferritic filler metal so that matching strength is obtained in the weldment and excellent toughness is retained even at 4K (-452°F) in both the weld metal and the heat-affected zone.
2. The promising properties obtained depend on two factors: the chemical cleanliness of the weldment as achieved in the GTAW process, and the microstructural refinement of the weldment and retention of microstructural refinement in the heat-affected zone, due to rapid thermal cycling of the weld region by the multipass GTA welding process.

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**References**


**WRC Bulletin 283**

**February, 1983**

**A Critical Evaluation of Fatigue Crack Growth Measurement Techniques for Elevated Temperature Applications**

by A. E. Carden

The report contains a discussion and evaluation of several crack length measurement techniques at elevated temperature and presents results from the experimental technique developed at the University of Alabama.

*Publication of this report was sponsored by the Subcommittee on Cyclic and Creep Behavior of Components of the Pressure Vessel Research Committee of the Welding Research Council.*

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