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HELIUM-INDUCED WELD DEGRADATION OF HT-9 STEEL

REFERENCE: Wang, C. A., Lin, H. T., Grossbeck, M. L., and Chin, B. A., "Helium-Induced Weld Degradation of HT-9 Steel," Effects of Radiation on Materials: 16th International Symposium, ASTM STP 1175, Arvind S. Kumar, David S. Gelles, Randy K. Nanstad, and Edward A. Little, Editors., American Society for Testing and Materials, Philadelphia, 1993.

ABSTRACT: Helium-bearing Sandvik HT-9 ferritic steel was tested for weldability to simulate the welding of structural components of a fusion reactor after irradiation. Helium was introduced into HT-9 steel to 0.3 and 1 atomic parts per million (appm) by tritium doping and decay. Autogenous single pass full penetration welds were produced using the gas tungsten arc (GTA) welding process under laterally constrained conditions. Macroscopic examination showed no sign of any weld defect in HT-9 steel containing 0.3 appm helium. However, intergranular micro cracks were observed in the HAZ of HT-9 steel containing 1 appm helium. The microcracking was attributed to helium bubble growth at grain boundaries under the influence of high stresses and temperatures that were present during welding. Mechanical test results showed that both yield strength (YS) and ultimate tensile strength (UTS) decreased with increasing temperature, while the total elongation increased with increasing temperature for all control and helium-bearing HT-9 steels.

KEY WORDS: Sandvik HT-9 steel, tritium doping and decay, GTA welding, HAZ cracking, helium bubble growth

Exposure of structural components of nuclear reactors to neutron irradiation will produce significant physical damage to materials and also result in the generation of helium by transmutation reactions [1-4]. Due to its extremely low solubility in metals, helium will diffuse and agglomerate to form helium bubbles after being trapped at point defects, dislocations and grain boundaries. The resulting degradation of structural components plays a decisive role in determining the useful

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lifetime of both fusion and fission nuclear reactors. Therefore, it is anticipated that repair and replacement of these degraded components by welding techniques will be required.

It has been known for several decades that the degradation of irradiated materials at elevated temperatures is partially exacerbated by the presence of helium [3]. At high temperature, helium bubbles grow rapidly under the influence of both stress and temperature, which typically occur in welding. As these bubbles grow along the grain boundaries, intergranular fracture can occur prematurely.

Welding of neutron-irradiated type 304 stainless steels has been reported on several occasions with varying degrees of heat-affected-zone (HAZ) cracking [5-8]. HAZ cracking has also been observed during gas tungsten arc (GTA) welding under fully constrained conditions of type 316 stainless steel, a candidate material for fusion reactor structural applications, containing greater than 2.5 appm helium [9-11]. Cracking was found to be intergranular and to run parallel to the welding direction. Underbead cracking also has been found in type 304 stainless steel containing 2.7 appm He using shallow penetration gas metal arc (GMA) overlay welding techniques [12-13]. The fracture facets of both HAZ and underbead cracking are decorated with dimples, which are attributed to the helium bubble growth at grain boundaries. In this study, Sandvik HT-9 ferritic steel, another type of candidate material for fusion reactor structural components, containing 0.3 and 1 appm helium, is investigated for weldability. In order to avoid hot-cell investigations, which are both lengthy and expensive, welding was performed on helium-doped steels to simulate the joining of irradiated materials.

EXPERIMENTAL PROCEDURES

The material investigated in this study was 0.76 mm thick HT-9 ferritic steel. The plates were solution annealed at 1050°C for one half hour in an inert gas atmosphere and then air-cooled. After annealing, the materials were tempered at 700°C for 1 hour in an inert gas atmosphere followed by air-cooling. To avoid hot-cell operations, helium was implanted into the steel using tritium doping and decay*. The steel was exposed to tritium gas at 1.7 and 38 MPa respectively at 300°C for 30 days. The dissolved tritium was allowed to decay to helium only during this charging period, and the excess tritium was then pumped off under a vacuum (10^{-3} Pa) at 400°C to prevent further helium generation. The resulting helium concentrations were 0.3 and 1 appm, respectively.

Autogenous single pass GTA welds were produced under laterally constrained conditions (the edges of the plates parallel to the weld direction were securely fastened to the welding table) to simulate the restraints encountered in practical weld repair of structural components. Figure 1 shows the welding stage used in this study. Welding was performed at 10 VDC and 28 A with a torch travel speed of 3.6 mm/s under a protective argon atmosphere. The resulting heat input was 77.8 J/mm, which produced a full penetration weld about 3 mm wide.

*The tritium doping was performed at Sandia National Laboratory, Livermore, CA.

FIG. 1--Laterally full constrained welding stage.

Optical microscopy was used to examine the microstructures of the steel across the weld. Diamond pyramid microhardness (DPH) was measured transverse to the weld to evaluate the microstructure-property relationship. In order to study mechanical properties of the welds, sheet tensile tests (gauge section of 12.7 mm long by 3.175 mm wide by 0.76 mm thick) were performed perpendicular to the welding direction with a strain rate of $5 \times 10^{-4} \text{ s}^{-1}$. Tests were conducted at temperatures between 25 and 700°C in a vacuum of $4 \times 10^{-5} \text{ Pa}$. Scanning electron microscopy (SEM) was also conducted to evaluate the weld integrity and the fracture surfaces.

RESULTS AND DISCUSSION

No visible cracking was observed in both control and 0.3 appm He welds. However, heat-affected-zone intergranular cracking was found in the steel containing 1 appm He. Cracks tended to run parallel to the welding direction, similar to the HAZ cracking observed in helium-bearing type 316 stainless steels [9-11]. Figure 2 shows the HAZ cracking in the as-welded HT-9 steel containing 1 appm helium. The fracture surface facets were decorated with dimples, which were attributed to bubble growth at grain boundaries. The shear ligaments separating the dimples have been rounded by surface diffusion, indicating that the cracking occurred at high temperature. However, cracking was observed only in the heat-affected zone within about 200 μm of the fusion/HAZ boundary and was limited to the beginning region of the weld. The cracking was found to occur along prior-austenite grain boundaries and was fully intergranular.

During the welding process helium bubbles, which were formed during the tritium pump-off period, will grow rapidly under the

FIG. 2--HAZ microcracking of HT-9 steel containing 1 appm He. (a) intergranular fracture along prior-austenite grain boundaries, and (b) grain boundary facets decorated with dimples.

influence of high temperatures and tensile stresses [9-11]. As the weld metal cools, shrinkage-induced tensile stresses are generated perpendicular to the weld direction and are enhanced by lateral constraint. Significant stress-assisted helium bubble growth occurs at grain boundaries perpendicular to the stress direction (parallel to the weld). Cracking occurs as the grain boundary strength, reduced by the growing bubbles, can no longer bear the shrinkage-induced tensile stress. It has been reported that high temperature alone is not sufficient to cause helium-induced HAZ cracking, but the combined action of high temperature and tensile stress is required [9-11].

Optical micrographs taken across HT-9 steel welds containing 0.3 and 1 appm helium are shown in Figures 3 and 4. In the fusion zone, the structure exhibited a mixture of untempered martensite and delta

FIG. 3--Optical microstructures of HT-9 steel, containing 0.3 appm He, across the weld region. (a) fusion zone, (b)-(e) sub-zones of HAZ, and (f) base metal. No visible cracking was found

FIG. 4--Optical microstructures of HT-9 steel, containing 1 appm He, across the weld region. (a) fusion zone, (b)-(e) sub-zones of HAZ, and (f) base metal. Intergranular cracking occurred in the two-phase region of the HAZ.

ferrite. The microstructures in the HAZ have been identified and divided into four distinct regions: two-phase, coarse-grained, fine-grained and over-tempered regions. The two-phase region adjacent to the fusion boundary also consists of untempered martensite and delta ferrite. The coarse-grained region is fully martensitic with a coarse prior-austenite grain size with a high hardness. The fine-grained region is martensitic with a fine prior-austenite grain size and is interspersed with undissolved carbides. The over-tempered region is a narrow softened band which was heated just below A_{c1} (about 780°C), and was typified by coarse martensite laths and carbides. The base metal microstructure consists of a mixture of lath-type martensite and carbide. The metallographic results again revealed that both control and 0.3 appm He steel welds were free of weld cracking, while the welds containing 1 appm helium revealed intergranular fracture along prior-austenite grain boundaries in the two-phase regions as shown in Figure 4b.

Following metallographic examination, 1000 gram load diamond pyramid microhardness (DPH) measurements were made across the weld region at room temperature. Figure 5 shows the microhardness traverses across the weld regions of control and helium containing materials. The results showed only minor variations in hardness between controls and He-doped specimens in both the fusion and heat-affected zones; the microhardness was insensitive to helium concentration at these levels. Despite the presence of the softer ferrite phase, the fusion zone exhibited nearly full martensite hardness. The fine-grained region exhibited a reduction in hardness relative to the hardness in the coarse-grained region, due to less carbon dissolved in this region. In addition, the over-tempered region showed a slightly lower hardness than the base metal. The microhardness showed a rapid drop between the fine-grained and over-tempered regions. This is due to the transition of the microstructure from untempered to tempered martensite.

Tensile test results of control and helium-doped specimens in unwelded and welded conditions of HT-9 steel are shown in Table 1. All the welded specimens failed in the base metal except for two specimens (one control and one 0.3 appm) that failed in the heat-affected zone. These two specimens showed much lower total elongation than the other specimens. Results from the tensile tests showed that the yield strength of HT-9 steel remained practically the same for all specimens, including control and helium-doped materials, in the welded and unwelded conditions (Figure 6). The similarity of welded and unwelded specimens is due to the fact that most of materials failed in the base metal. The result is also consistent with the observation in microhardness measurements that there is little strengthening associated with the presence of helium at 0.3 and 1 appm concentration. Similar to the results of yield strength, the ultimate tensile strength of HT-9 decreased with increasing temperatures for control and helium-doped materials as shown in Figure 7. Although there is some variation, the ultimate tensile strength of helium-doped materials was about the same level as that of controls and all welded materials at the temperatures tested.

Results of total elongation as a function of test temperatures is given in Figure 8. In general, the total elongation increased with increasing test temperature for HT-9 steels. In contrast, the total

FIG. 5--Diamond pyramid microhardness (DPH) traverses across the weld region of HT-9 steel. No strengthening effect was observed at both helium concentration levels.

FIG. 6--Temperature and helium concentration dependence of yield strength of HT-9 steel. YS decreases with increasing temperature.

TABLE 1--Tensile test data of HT-9 steels

Material	Temp. (°C)	UTS (MPa)	YS (MPa)	Elong. (%)
<u>Unwelded</u>				
Control	20	1001.2	835.0	4.7
	20	1013.8	866.9	5.4
	500	869.5	777.0	3.0
	500	845.3	760.8	3.2
	600	654.0	616.0	7.4
	600	670.3	622.4	7.0
	700	425.5	414.2	10.5
	700	447.8	422.8	12.2
0.3 appm	20	1034.5	903.3	6.1
	500	760.8	695.1	4.5
	600	653.4	604.0	5.5
	700	409.0	399.7	9.1
1.0 appm	20	1038.0	983.8	5.1
	500	855.6	744.1	5.0
	600	702.4	681.3	6.6
	700	426.2	416.7	10.6
<u>Welded</u>				
Control ^a	20	983.0	895.4	5.5
	20 ^b	965.3	890.1	0.7
	500	815.6	765.8	4.2
	500	739.4	708.6	1.8
	600	756.4	701.6	3.4
	600	713.0	676.0	3.8
	700	418.2	408.2	7.2
	700	475.7	464.5	6.2
0.3 appm ^a	20 ^b	961.0	827.0	0.3
	500	795.7	736.4	1.6
	600	688.6	633.4	3.9
	700	372.6	356.8	8.7
1.0 appm ^a	20	992.6	882.3	3.1
	500	812.1	764.4	2.6
	600	696.2	675.2	5.3
	700	410.4	400.8	8.0

^aSpecimens failed in the base metal.^bSpecimens failed in the heat-affected zone.

FIG. 7--Temperature and helium concentration dependence of ultimate tensile strength of HT-9 steel. UTS decreases with increasing temperature.

FIG. 8--Temperature and helium concentration dependence of total elongation of HT-9 steel. Total elongation increases with increasing temperature.

FIG. 9--Fractography of the unwelded HT-9 steels, control and helium-doped, after tensile tests at 20 and 700°C. Dimple size increases with increasing test temperature.

FIG. 10--Fractography of the welded HT-9 steels, control and helium-doped, after tensile tests at 20 and 700°C. Dimple size increases with increasing test temperature.

elongation decreased with increasing temperature in the case of type 316 stainless steel containing 0.18 to 256 appm helium [9-11]. This result suggests that total elongation of HT-9 steel is insensitive to the presence of helium at concentrations of 0.3 and 1 appm. Furthermore, results also showed that the unwelded HT-9 specimens generally exhibited greater total elongation than welded HT-9 specimens. This is due to the fact that the higher strength of weld metal will give less plastic deformation than that of base metal.

Figure 9 shows the fracture surfaces of the unwelded, control and helium-doped HT-9 steels after tensile testing at 20 and 700°C. Generally, all the specimens revealed dimple rupture along lath boundaries. The size of dimples tended to increase with increasing temperature, which is attributed to increased helium bubble growth with increasing temperature. The fracture surface features of welded control and helium-doped HT-9 steels are shown in Figure 10. The fracture characteristics are similar to those of unwelded specimens. All specimens showed dimple rupture, except for the welded specimen with 0.3 appm He tested at 20°C, which failed predominantly by brittle cleavage in the HAZ due to the presence of untempered martensite.

CONCLUSIONS

The following conclusions were drawn from the present study:

1. Heat-affected zone intergranular cracking occurred during GTA welding of HT-9 ferritic steel containing 1 appm helium.
2. Heat-affected-zone cracking was a result of helium bubble growth at grain boundaries under the influence of high temperatures and stresses during welding.
3. No strengthening effect on HT-9 steel was observed at helium concentrations of 0.3 and 1 appm.
4. Both YS and UTS decreased with increasing temperature while total elongation increased with increasing temperature.

ACKNOWLEDGMENT

Research Sponsored by the Office of Fusion Energy, U.S. Department of Energy, under Contract No. DE-AC05-84OR21400 with Martin Marietta Energy System, Inc., Grant No. DEFG0585ER52139C with Auburn University, and the Division of Materials Science, U.S. Department of Energy, under the SHaRE program under Contract No. DE-AC05-76OR00033 with Oak Ridge Associated Universities.

REFERENCES

- [1] H. S. Rosenbaum, "Microstructures of Irradiated Materials," Treaties on Materials Science and Technology, Vol. 7, Academic Press, Inc., New York (1975).
- [2] E. E. Bloom, "Mechanical Properties of Materials in Fusion Reactor First-Wall and Blanket Systems", Journal of Nuclear Materials, 85&86(1979)795-804.

- [3] R. S. Barnes, "Embrittlement of Stainless Steel and Nickel-Based Alloys at High Temperature Induced by Neutron Irradiation," Nature, 206(1965)1307-1310.
- [4] D. R. Harries, "Neutron Irradiation-Induced Embrittlement in Type 316 and Other Austenitic steels and alloys," Journal of Nuclear Materials, 82(1979)635-641.
- [5] J. P. Maloney, et al., "Repair of a Nuclear Reactor Vessel", DP-1199, E. I. du Pont de Nemours & Co., Savannah River Laboratory, Aiken, SC, June 1969.
- [6] M. M. Hall, Jr., A. G. Hins, J. R. Summers, and D. E. Walker, "Fusion Welding of Irradiated AISI 304L Stainless Steel Tubing", Weldment: Physical Metallurgy and Failure Phenomena, Proceedings of the Fifth Bolton Landing Conference, August 1978, pp. 365-378.
- [7] S. D. Atkin, "Weld Bend Tests on Irradiated, 20% Cold-Work 316 Stainless Steel", Fusion Reactor Materials, ADIP Semiannual Progress Report, Sept. 30, 1981, DOE/ER-0045/7, pp. 110-117, U.S. DOE, Office of Fusion Energy.
- [8] W. K. Kanne, C. L. Angerman and B. J. Eberhard, "Weldability of Tritium-Charged 304L Stainless Steel", DP-1740, E. I. du Pont de Nemours & Co., Savannah River Laboratory, Aiken, SC, February 1987.
- [9] H. T. Lin, S. H. Goods, M. L. Grossbeck, and B. A. Chin, "Helium-Induced Degradation in the Weldability of an Austenitic Stainless Steel", Effect of Radiation on Materials, 14th International Symposium, Volume I, ASTM STP 1064, N. H. Packan, R. E. Stoller, and A. S. Kumar, Eds., American Society for Testing and Materials, Philadelphia, 1989, pp. 301-314.
- [10] H. T. Lin, "Simulation of the Welding of Irradiated Materials", ORNL/TM-11158, Oak Ridge National Laboratory, Oak Ridge, TN, 1989.
- [11] H. T. Lin, M. L. Grossbeck and B. A. Chin, "Cavity Microstructure and Kinetics During Gas Tungsten Arc Welding of Helium Containing Stainless Steel," Metallurgical Transactions A, V21A, September 1990, pp.2585-2596.
- [12] S. H. Goods and C. W. Karfs, "Helium-Induced Weld Cracking in Low Heat Input GMA Weld Overlays," Welding Journal, Vol. 70, No. 5, 1991, pp. 123-132.
- [13] W. R. Kanne, Jr., D. A. Lohmeier, L. R. Louthan, Jr., D. T. Rankin, E. A. Franco-Ferreira, G. J. Bruck, A. Madeyski, R. P. Shogan and G. G. Lessmann, "Weld Repair of Helium Degraded Reactor Vessel Material", Proceedings of the Fifth International Symposium on Environmental Degradation of Materials in Nuclear Power System - Water Reactors, August 25-29, 1991, Monterey, CA, pp. 390-402, American Nuclear Society.

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