RECENT PROGRESS ON GAS TUNGSTEN ARC WELDING OF VANADIUM ALLOYS — J. F. King, M. L. Grossbeck, G. M. Goodwin, and D. J. Alexander (Oak Ridge National Laboratory)

SUMMARY

This is a progress report on a continuing research project to acquire a fundamental understanding of the metallurgical processes in the welding of vanadium alloys. It also has the goal of developing techniques for welding structural vanadium alloys. The alloy V-4Cr-4Ti is used as a representative alloy of the group; it is also the prime candidate vanadium alloy for the U.S. Fusion Program at the present time. However, other alloys of this class were used in the research as necessary. The present work focuses on recent findings of hydrogen embrittlement found in vanadium alloy welds. It was concluded that the atmosphere in the inert gas glove box was insufficient for welding 6mm thick vanadium alloy plates.

PROGRESS AND STATUS

Introduction

The vanadium alloy welding program at ORNL focuses on both gas tungsten arc (GTA) and electron beam (EB) welding methods. The alloy V-4Cr-4Ti of Teledyne Wah Chang heat 832665 was used for the welding research. However, some welds were made on V-45Cr-5Ti from Teledyne Wah Chang heat 832394. Thin plate, usually 0.76 or 1.0 mm in thickness is used for welds that will be studied by microanalytical techniques or by tensile testing. Plate 6mm in thickness is used for welds to be studied by charpy testing as well as microanalytical techniques. The 6mm plate proved to be more challenging since multiple passes are required and the larger heat sink results in slower cooling of the weld zone with consequent contamination by higher levels of impurities.

Experimental Methods and Results

GTA welds in the 6mm plate were made in an argon filled glove box previously evacuated to a pressure of the order of 10-4 Pa. Moisture levels typically below 40 ppm as measured by a CEC moisture monitor were achieved. A 75° included angle V-groove with a 2.4 mm root opening joint geometry was used for the weldments. The filler wire was of matching composition of the same V-4Cr-4Ti heat as the base metal. Multi-pass welds were made using direct current, electrode negative, at a current range of 100 to 140 amperes and 12 volts.

Earlier research on this alloy and the similar V-5Cr-5Ti alloy using similar welding methods has shown that a post-weld heat treatment (PWHT) is necessary [1]. Charpy V-notch testing demonstrated shifts in the ductile to brittle transition temperature (DBTT) of greater than 200°C following welding. Post weld heat treatments resulted in some cases, depending upon the alloy, initial heat treatment of the base metal, and the temperature of the PWHT, in a return of the DBTT to a lower temperature than that of the base metal. This phenomenon was previously explained in terms of precipitation of oxygen, primarily with titanium. As has been observed in refractory metals, precipitates dissolve upon melting of the alloy [2]. The rapid cooling during welding quenches large concentrations of oxygen into the matrix leading to severe hardening and embrittlement. The PWHT allows sufficient time for the oxygen and titanium to form TiO and similar titanium oxycarbonitrides which getter the oxygen and produce a ductile matrix.

Data from Charpy testing is shown in Fig. 1 for as-welded specimens as well as specimens given post-weld heat treatments at various temperatures for a fixed time of two hours. All data were obtained from a single weld, designated GTA10. More consistent curves were usually obtained with other welds; however, occasionally a datum point would fall on the lower shelf with adjacent points at both higher and lower temperatures on the upper shelf. In these cases, the fracture surfaces of the lower-shelf samples revealed almost entirely cleavage fracture. This report will address the problems encountered with consistency of the data and the possible sources of the observed embrittlement. In Fig. 1, the trends are discernable, but the scatter caused by erratic lower shelf values is so high that it is difficult to draw conclusions from the plot. As with previous specimens, the outlying lower shelf specimens failed by cleavage. They did not fail due to an existing flaw in the weld resulting in a reduced load support area. The cleavage fracture resulted from a serious embrittlement mechanism.

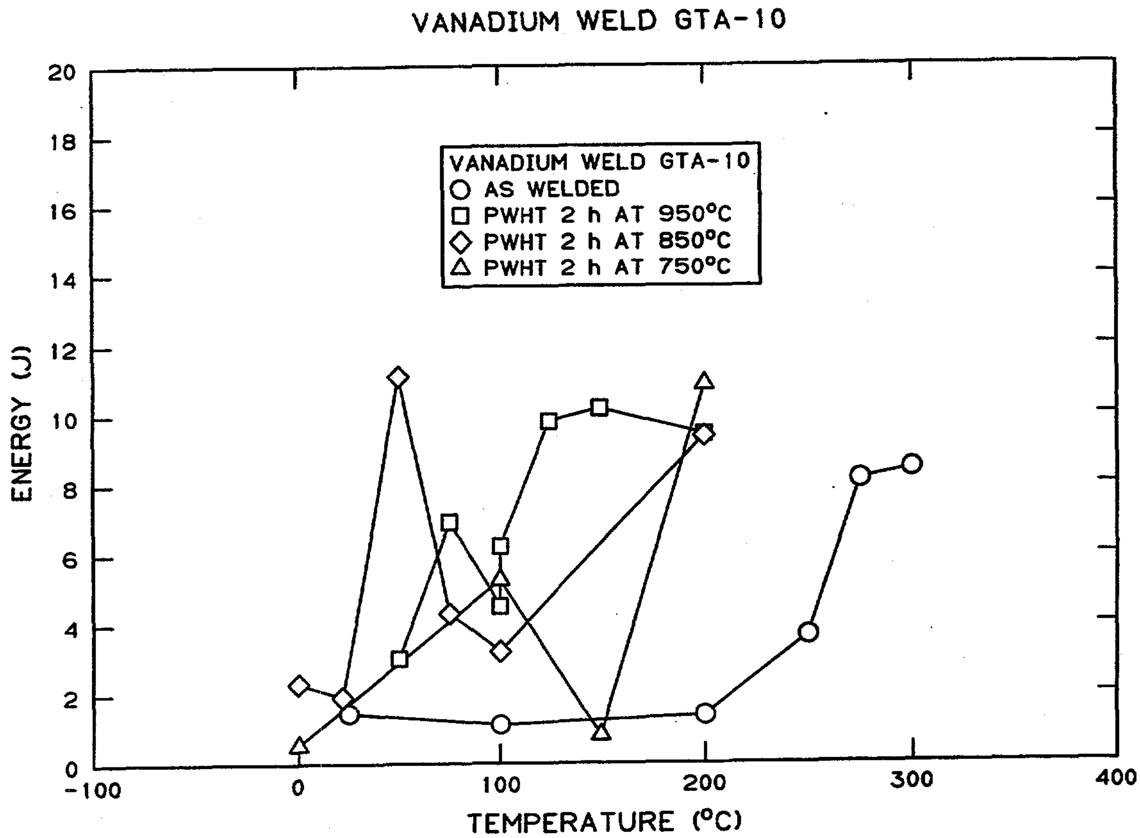

Fig. 1. Charpy impact data from gas-tungsten-arc weld GTA-10 in V-4Cr-4Ti. Samples were given three different post-weld heat treatments (PWHT) as indicated.

The problem was more emphatically brought to light with the subsequent GTA weld (Designated GTA11). GTA11 was prepared using the same procedure as for the previous weld, GTA10. When a sample was cut from one end of the weld for metallography, the plate cracked. As can be seen in Fig. 2, brittle transgranular cracks appear. There are also intergranular cracks connecting with transgranular cracks, but it is not apparent which initiated the failure. The series of non-linking parallel cleavage cracks is very similar to those observed in niobium embrittled by hydrogen [3].

DISCUSSION

Because of the similarity with hydrogen embrittlement in this group of metals, analyses were made for hydrogen and other interstitials. The analyses were done by inert gas fusion analysis by the Leco Corporation, St. Joseph, MI. Samples of as received alloy as well as acid -etched material was analyzed with the results shown in Table 1. Analyses from earlier welds are included in the table to aid in locating the source of the embrittlement. The hydrogen concentration in the etched V-4Cr-4Ti sample shows that the acid etch does not appear to add hydrogen to the alloy. In general, EB welding does not suffer from the hydrogen contamination problem except for one weld, EBW11. It is believed that a leak in the system or a

nother failure in the vacuum system is responsible for the elevated hydrogen concentration in EBW11; the hydrogen concentration is four to ten times that observed in two earlier EB welds. The GTA welds appear to have higher hydrogen concentrations than the EB weld samples. The weld of present interest is GTA11 which had hydrogen concentrations between 50 and 60 wt. ppm in both the fusion zone and the base metal. Considering the high diffusivity of hydrogen in vanadium, the similarity of hydrogen concentrations in the fusion zone and base metal is not unexpected.

Table 1. Concentration of Interstitial Elements, wt. ppm

Base Metal	Weld Metal	H	O	N	C
	GTA11	58.5	446	288	—
GTA11		53.5	364	96	—
V-4Cr-4Ti (as received)		316	114	—	—
V-4Cr-4Ti (acid etched)		1.2	—	—	—
Weld Wire V-4Cr-4Ti		—	360	109	155
	EBW11	36.4	327	99	—
EBW11		23.1	323	97	—
	GTA2 (VQ11 ductile)	11.2	—	—	—
	GTA2 (VQ09 brittle)	10.1	—	—	—
	GTA3	20.5			
	GTA8		410	98	—
GTA8			332	96	—
	EB2	2.8	—	—	—
	EBW12	7.9			

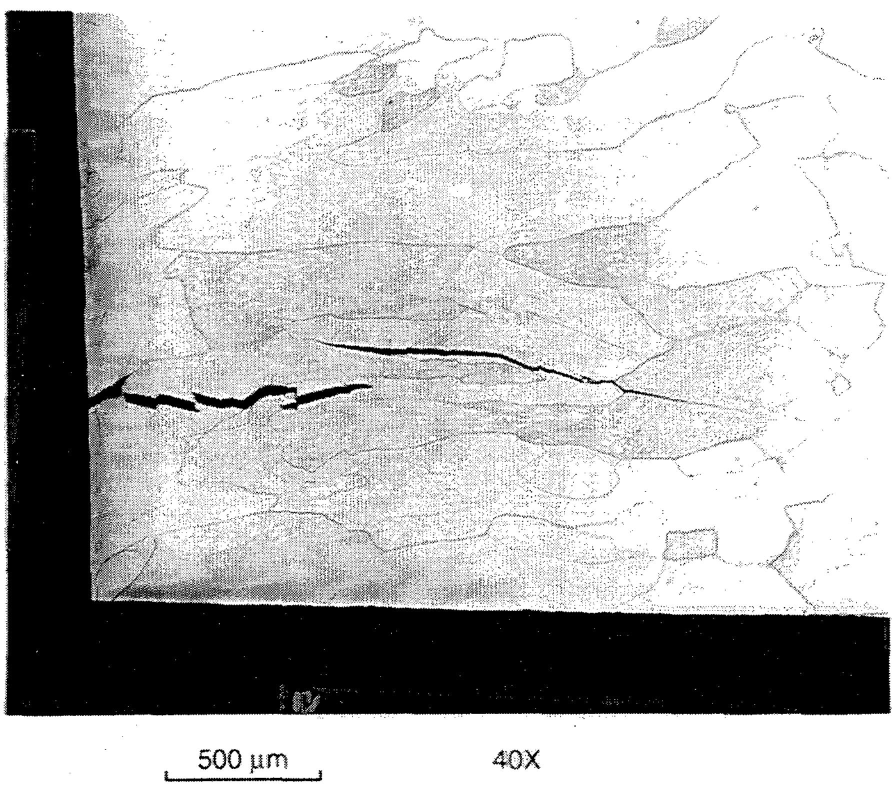

Fig. 2. View of fusion zone of weld GTA-10 in V-4Cr-4Ti showing brittle cracking following cutting of metallography sample (sample 96-0218).

Eustice and Carlson have observed only a mild drop in ductility at room temperature upon the addition of 50 wt. ppm hydrogen [4]. However, in their research, the onset of embrittlement occurs at slightly lower temperatures. Westlake has determined the hydride solvus temperature to be -68°C at 50 wt. ppm hydrogen in vanadium [5]. These observations do not support embrittlement and cracking at such a level of hydrogen. However, hydrogen embrittlement has been shown to occur at slightly higher temperatures in refractory metals because of stress-induced hydride at the crack tip [3]. Such a shift is only expected to be on the order of 10-20°C; however, although 50 wt. ppm does not appear to be an embrittling concentration, the stronger an alloy becomes, the larger the shift in the formation of hydride. As shown in Table 1 the oxygen and nitrigen concentrations in GTA 11 were also high. Such a high impurity concentrations, which might have resulted from a leak in the welding chamber (although none were found), could harden the alloy thus increasing the shift in the hydride solvus beyond the expected 10 or 20 degrees. The chromium and titanium alloying elements could also shift the phase diagram sufficiently to make the embrittlement occur at a slightly higher temperature. Experimental evidence exits on the V-4Cr-4Ti alloy itself. Roehrig et. al observed that 500 wt. ppm hydrogen was necessary to induce room temperature embrittlement as measured by tensile testing but that, in the presence of 850 wt. ppm oxygen, a factor of four reduction in ductility was observed with only 90 wt. ppm hydrogen [6]. Another synergistic mechanism might exist in addition to lattice hardening. In any case, it would be expected to be operational in the welded sample as well. Detailed x-ray phase analysis would have to be done to confirm this mechanism.

At the present time, a better approach is to reduce moisture and oxygen contamination in the welding glove box atmosphere to achieve lower levels of interstitials in the welds. An inert gas purification system as shown in Fig. 3 has been added to the welding glove box. The purification system consists of a gettering furnace which contains a high purity titanium charge that removes oxygen, nitrogen, carbon, and water to extremely low levels. A molecular sieve trap will also be added at the gettering furnace inlet to reduce moisture levels still further. This system is undergoing testing and adjustments at the present time.

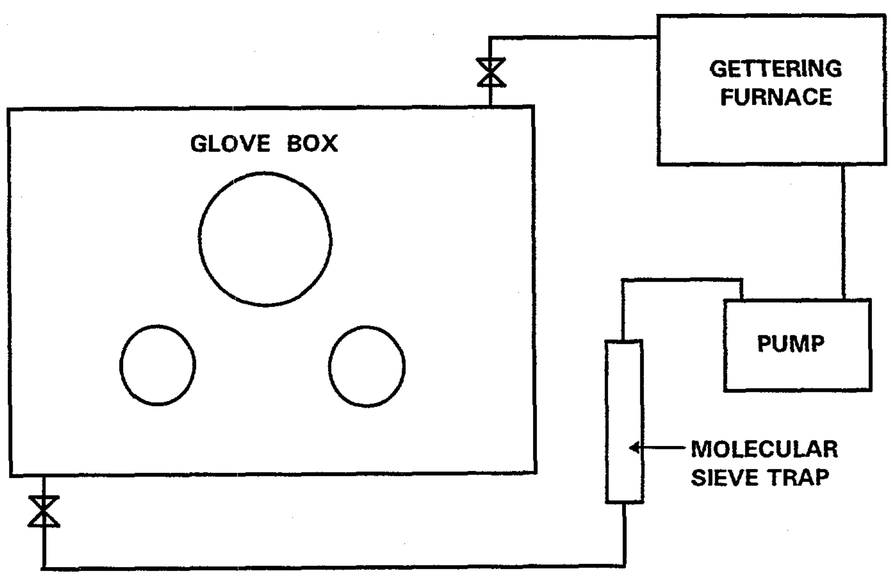

Fig. 3. Diagram of purification system to remove oxygen and water from welding chamber.

Conclusions

1. A post-weld heat treatment is required for welding V-Cr-Ti alloys.
2. Special precautions must be taken to avoid hydrogen embrittlement in GTA welding of V-Cr-Ti alloys.

Future Work

This is a continuing research project. The gettering system is expected to be put into operation during the next reporting period. This will enable uncontaminated welds to be made thus permitting detailed studies of the effects of post-weld heat treatments. It will also permit studies of the effects of impurities in welds.

REFERENCES

1. J. F. King, G. M. Goodwin, and D. J. Alexander, "Welding Development for V-Cr-Ti Alloys," *Fusion Materials Semiannual Progress Report for Period Ending September 30, 1994,* DOE/ER-0313/17 (1995), p. 152.

2. J. R. DiStefano, *Review of Oxidation of Nb-1Zr,* Oak Ridge National Laboratory, ORNL/TM-10907 (1989).

3. M. L. Grossbeck and H. K. Birnbaum, "Low Temperature Hydrogen Embrittlement of Niobium II--Microscopic Observations," *Acta Met. 25* (1977), p. 135.

4. A. L. Eustice and O. N. Carlson, "Effect of Hydrogen on the Tensile Properties of Iodide Vanadium," *Trans. AIME 221* (1961), 238.

5. D. G. Westlake, "A Resistometric Study of Phase Equilibria at Low Temperatures in the Vanadium-Hydrogen System," *Trans AIME 239* (1967), p. 1341.

6. H. D. Roehrig, private communication, Oak Ridge National Laboratory, 1996.