Progress Toward a More Weldable A-286

In this first of two reports, the mechanism and the elements responsible for weld metal hot cracking are identified

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ABSTRACT. This paper reports the results of a study concerned with high temperature phase equilibria in A-286 stainless steel and the role of the individual alloying elements on fusion zone (weld metal) cracking of A-286. Results also indicated that the composition of the high temperature grain boundary phase thought to be responsible for HAZ microfissuring in A-286 is more complex than that reported in the literature, and that a more weldable alloy can be achieved by minor adjustments in chemistry.

Introduction

A-286, an age hardenable austenitic stainless steel, has for a long time been recognized as an alloy highly susceptible to weld hot cracking. The alloy suffers from both heat-affected zone (HAZ) microfissuring and fusion zone (weld metal) hot cracking.

Several hot cracking mechanisms have been suggested to explain weld HAZ cracking in A-286 (Refs. 1-3). The most prevalent evidence indicates that a low melting point eutectic in the grain boundaries is associated with the hot cracking. Vagi and Martin (Ref. 1) observed intergranular cracks associated with a secondary grain boundary phase in specimens heated to 1340 C. Through x-ray diffraction they identified this phase as being the Fe_2Ti Laves phase, which forms a eutectic with Fe at about 1300 C.

Blum and Witt (Ref. 2) also proposed that HAZ cracking is associated with the liquation of the grain boundaries by the Fe-Fe_2Ti eutectic which allows the boundaries to separate under the welding stresses.

High Temperature Phase Stability

The chemistries of A-286 heats studied in this program are listed in Table I. Heat 1 is a standard commercial heat of A-286 and heats 2-9 were special experimental heats.

Since grain boundary liquation has been suggested to occur, special heat treatments were made to produce a sufficient quantity of phase for qualitative analysis.

Samples of Heat 1 were heat treated at 1350 C for times of 5 min to 2 h. The amount of grain boundary phase increases with increasing time, but there were no indications that different phases form as a func-

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tion of time at 1350 C. The typical microstructure of samples heated at 1350 C for 2 h is shown in Fig. 1. Electron microprobe results, Fig. 2, show that the grain boundaries are highly enriched in Ni, Ti, Si, and C and depleted in Fe and Cr. The microprobe results show at least two different Ti phases in the boundaries, one heavily concentrated in Ti and appearing to be associated with C, probably TiC formed on cooling, and another phase enriched in Ti, Ni and Si. X-ray diffraction on particles dissolved from the matrix with a mixed acid etch showed two phases: a TiC type (a_o = 4.33) and a A_2B type Laves phase (a_o = 4.76, C_o = 7.71). The Laves phase reported to form in iron base austenitic alloys in the range between 1200-2000 F (649-1093 C) has the general formula $(Fe, Cr, Mn, Si)_2$ (Mo, Ti, Cb). The large amount of nickel we observed in the high temperature phase is not reported in this lower temperature phase. It has been reported that high Si stabilizes M_2Ti while Ni favors the formation of the M_3Ti type phase (Ref. 4). However, only the M_2Ti type phase was observed in the particles extracted in the mixed acid etch.

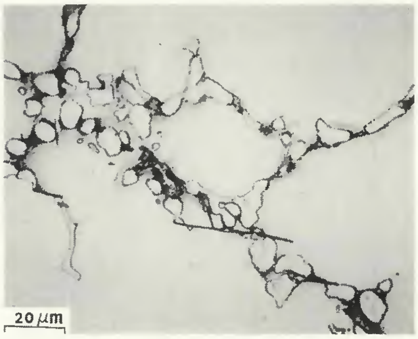
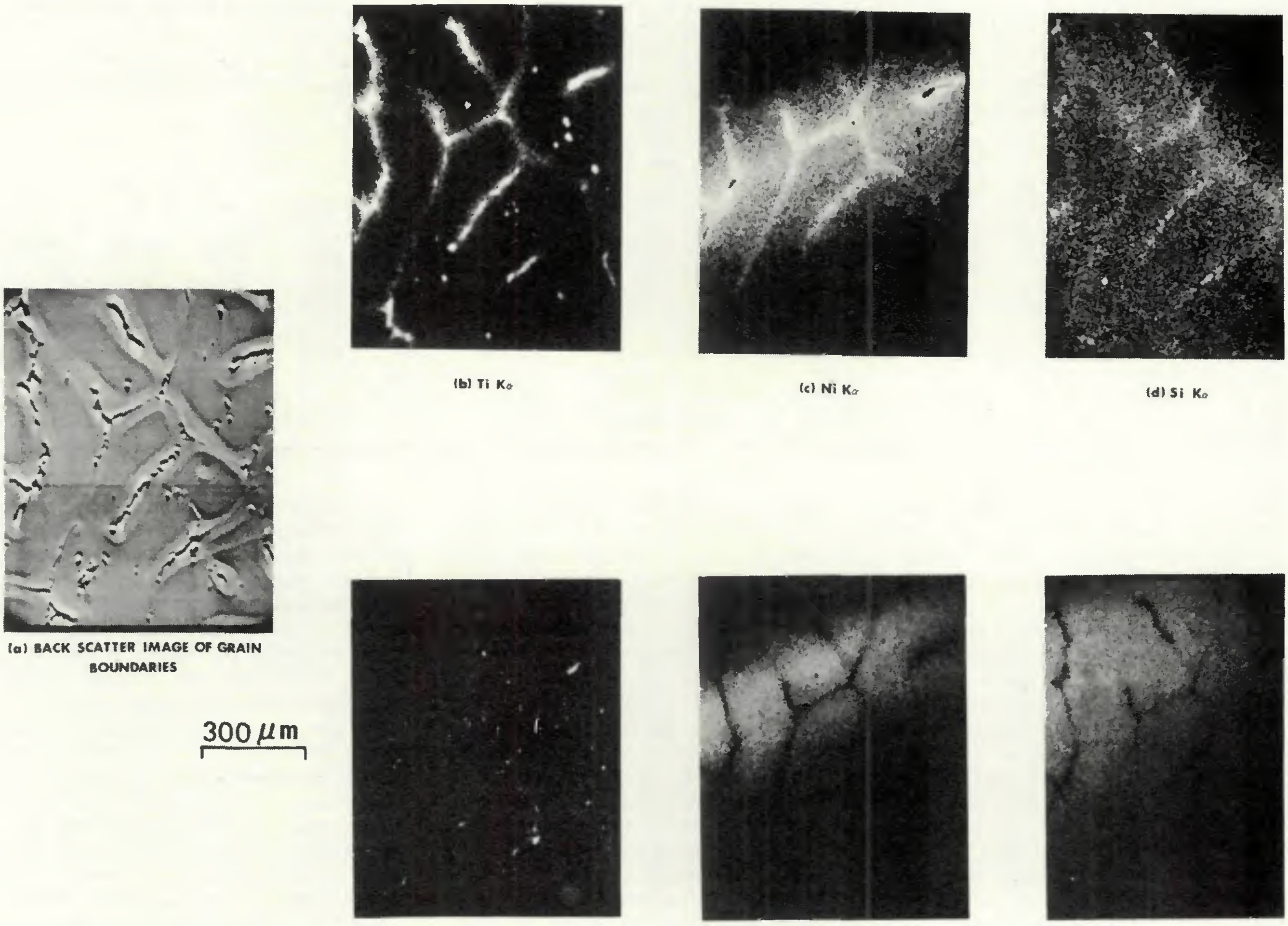


Fig. 1 — Sample heated at 1350 C for 2 h exhibiting large amounts of grain boundary phases

The large number of elements reported in solid solubility of the Laves phase and the additional observation in this work of large solubility of Ni and Si suggest that through chemistry modifications, the phase equilibrium and properties of this phase could be altered. By this approach it appeared possible to reduce weld hot cracking without eliminating the formation of the Laves phase.

Electron Beam Welding

Electron beam welding tests were conducted on Heats 2-9 in Table I to determine the effect of chemistry on weldability. Plates 3 × 3 × 0.25 in. were solution treated at 995 C (1825 F) for 45 min and water quenched. This gave an ASTM grain size of about 6. Five 90-100% penetration circular bead on plate welds with diameters of ½ to 2½ in. were made on each plate. Welding parameters of 75 ipm, 150 kV, 17 mA and sharp 8 in. focus, found to be very severe on nominal chemistries, were used. By using slower welding speeds with the parameters adjusted to give the same penetrations, cracking could be greatly reduced or eliminated. For these tests, however, the most severe conditions were used to discern more easily the relative hot cracking susceptibility of each heat. The above test, designed to maximize welding constraints, was used to study mainly fusion zone cracking using only a limited amount of material. After welding, the samples were sectioned and examined for gross



(a) BACK SCATTER IMAGE OF GRAIN BOUNDARIES

(b) Ti K_α

(c) Ni K_α

(d) Si K_α

(e) C K_α

(f) Fe K_α

(g) Cr K_α

Fig. 2 — X-ray microprobe results of sample shown in Fig. 1

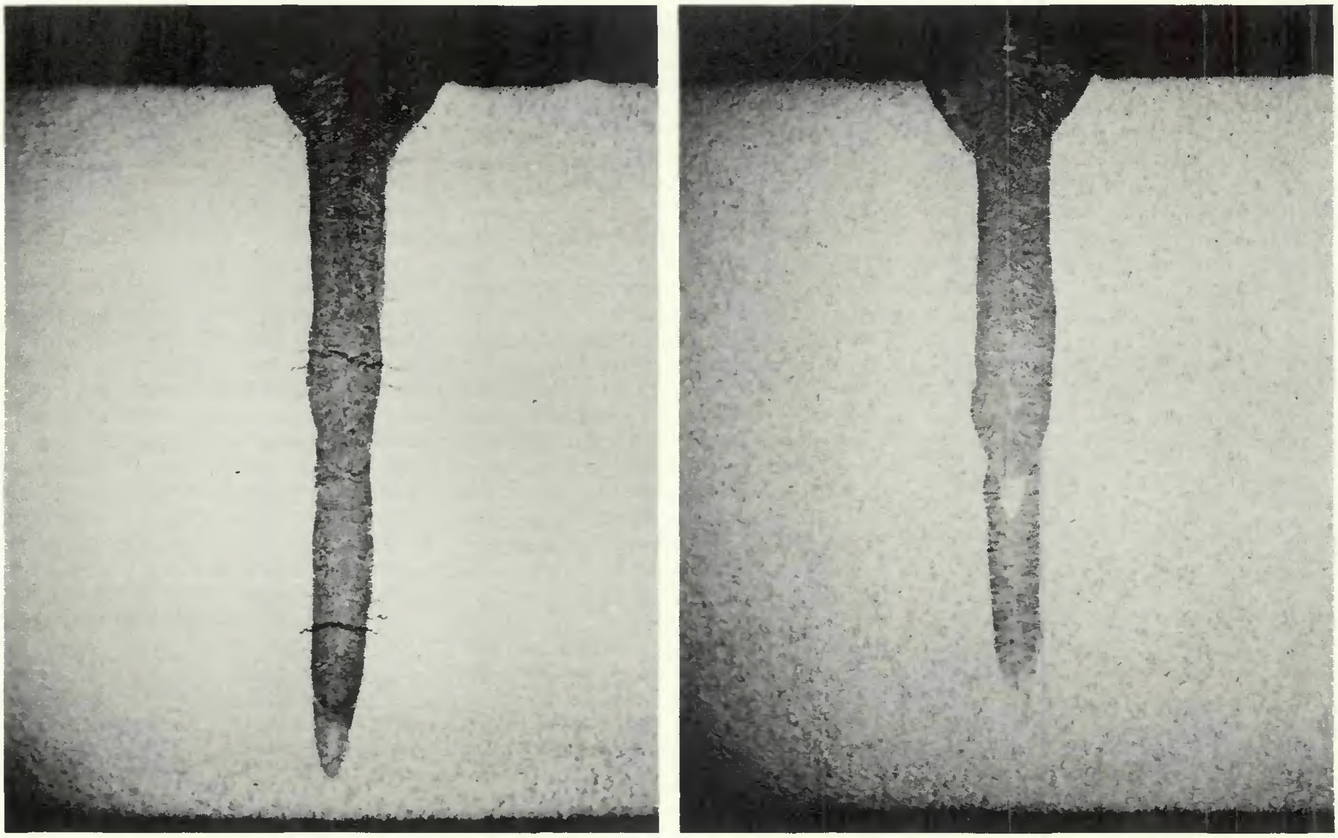


Fig. 3 — Partial penetration EB welds in 0.25 in. plate showing weld overlap. (Left) — Heat 2 exhibiting gross cracking. (Right) — Heat 6 free of gross cracks

Table 1 — A-286 Experimental Heats

Heat no.	Chemistry, wt. %, balance Fe C	Mn	P	S	Si	Cr	Ni	Al	Mo	V	Ti	B
1	0.037	1.22	0.015	0.012	0.70	14.29	25.59	0.22	1.37	0.22	2.25	0.0046
2	0.063	1.50	0.004	0.005	0.58	14.45	25.40	0.22	1.27	0.26	2.25	0.0073
3	0.031	1.51	0.004	0.006	0.63	14.50	25.70	0.24	1.27	0.26	2.22	0.0059
4	0.025	1.47	0.004	0.005	0.61	14.50	25.50	0.25	1.26	0.25	2.20	0.0009
5	0.028	0.13	0.006	0.005	0.25	14.77	26.01	0.23	1.25	0.26	2.18	0.0063
6	0.027	0.11	0.004	0.005	0.16	14.70	24.70	0.10	1.25	0.25	2.18	0.0014
7	0.028	1.54	0.004	0.005	0.63	14.25	34.70	0.25	1.24	0.25	2.15	0.0064
8	0.025	1.51	0.004	0.005	0.60	14.75	25.45	0.66	1.27	0.23	1.60	0.0012
9	0.025	1.39	0.010	0.008	0.58	15.20	23.50	0.24	1.25	0.25	2.63	0.006

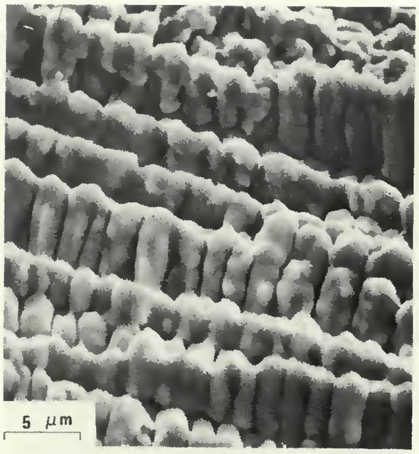
fusion zone cracking. These results are given in Table 2 where the percent of weld sections examined which exhibited gross fusion zone cracks, Fig. 3, are shown.

In Table 2, Heats 2 and 3 compare the effect of carbon modifications on weldability. As shown, the reduction of C from 0.06 to 0.03% in Heat 3 had no beneficial effect on fusion zone cracking. Both the increase in Ti from 2.25 to 2.63% in Heat 9 and the increase in Ni from 25 to 35% in Heat 7 had beneficial effects. In addition, Heats 4 and 5 showed that both B and Si are extremely detrimental to fusion zone hot cracking. Comparing the results from Heats 4 and 8, it appears that the substitution of about half a percent of Ti with Al was not effective in reducing fusion zone cracking in the electron beam welds. However, it is not certain this would be the case if boron had not been eliminated. No gross cracking was observed in Heat 6 in which boron, manganese and Si are low.

Discussion

A scanning electron microscope (SEM) was used to examine the type of fusion zone cracks shown in Fig. 3. The round, smooth appearance of the dendritic structure of the crack surface, Fig. 4, strongly suggests that the cracks were formed during the last stages of solidification. Figure 5 shows the segregation of the low melting phase and its association with the fusion zone cracks. By using the energy-dispersive x-ray spectrometer of the SEM it was found that this phase is also enriched in Ti, Si, Ni and possibly Mo and depleted in Fe and Cr with respect to the matrix. This segregated phase of the fusion zone appears to be the same phase that forms in the weld HAZ. In many cases this phase was seen to be continuous from the fusion zone into the weld HAZ. The SEM x-ray results suggested that the phase was very close to the same composition in both the fusion and HAZ and is probably the same grain boundary phase formed at high temperatures. That result is in agreement with the electron microprobe studies on the same material.

By knowing the qualitative composition of the phase responsible for



FUSION ZONE CRACK SURFACE

Fig. 4 — SEM micrograph of an electron beam weld fusion zone crack surface

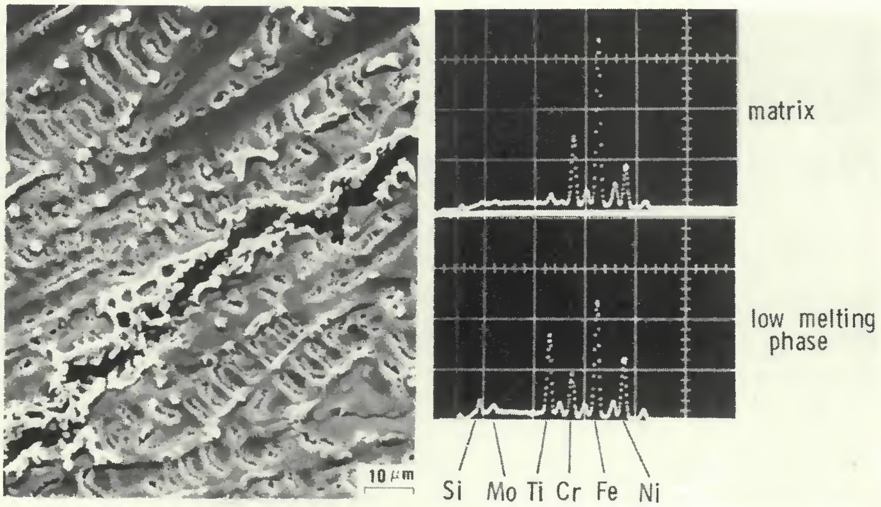


Fig. 5 — Fusion zone crack associated with the segregated low melting phase. SEM energy dispersive x-ray spectrometer results showing an enrichment of Ni, Ti, Si and Mo in the segregated phase

Table 2 — Electron Beam Weld Results

Heat no.	Major chemistry modification	% weld sections exhibiting gross cracks
2	Standard	72
3	Low C	70
7	Low C + 10% Ni	45
8	Low C, 1.6Ti, 0.66Al, No B	38
5	Low C, Mn, Si	18
9	Low C, 2.75 Ti	10
4	Low C, No B	8
6	Low C, Mn, Si, No B	0

fusion zone cracking the effect of alloy composition on cracking can be better understood.

Since the low melting segregated phase is enriched in Ti, the decrease in fusion zone cracking with increased Ti, Heat 9, and the apparent increase in cracking with decreased Ti of Heat 8 with respect to Heat 4, can be explained by the shrinkage brittleness theory (Ref. 5). This theory proposes that during the last stages of solidification there is a minimum amount of eutectic required to heal all the cracks formed, and if a critical, subminimum amount of eutectic forms, a maximum amount of cracking results. Since the eutectic formed during solidification is also enriched in Ni, the positive effect of the increased Ni can also be explained by this theory.

SEM results showed that Si is also associated with the segregated phase. The reduction of Si could then greatly affect either or both the quantity and properties, such as the melting point of this phase. Both B and Si form low melting point eutectics with Fe and Ni. Therefore, small amounts of either of these two elements could lower the eutectic melting points and thus greatly aggravate fusion zone cracking.

It is apparent from the electron beam welding results that increases in both Ni and Ti and decreases in both Si and B are extremely effective in reducing fusion zone cracking. Therefore, through chemistry modifications, the susceptibility of A-286 to fusion zone cracking can be greatly reduced. This is shown in Fig. 3 where gross cracking is observed in Heat 2, the standard low C chemistry, and no fusion zone cracking was observed in Heat 6, the low C, Mn, Si, no-B heat.

Since A-286 is known to suffer from HAZ microfissuring, it may be that some of the chemistry modifications which improve fusion zone cracking may be detrimental to HAZ cracking. Therefore, both Gleeble and spot varestraint tests are being conducted to determine the effect of chemistry modifications on HAZ microfissuring. These results will be reported in a future paper.

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