Effect of Alloy Modifications on HAZ Cracking of A-286 Stainless Steel

The mechanisms by which certain elements can produce beneficial or detrimental effects are studied by varestraint and hot ductility testing

BY J. A. BROOKS

ABSTRACT. This investigation examined the heat-affected zone (HAZ) cracking mechanisms in A-286 using spot varestraint and hot ductility testing techniques. The better understanding of these mechanisms made possible slight alloy modifications which greatly reduce HAZ cracking.

A partially melted HAZ exists in A-286 welds. Cracking was found to occur both in this partially melted region and further away from the fusion zone where no indication of melting was visible. Hot ductility data indicate that cracks can propagate both in the liquid state and after solidification of the heavily liquated region. The grain boundary liquation resulted from Laves phase formation nucleating largely around decomposing TiC-type carbides. The complex composition of the Laves phase allows the properties and quantity of the melt phase to be significantly affected by slight chemistry modifications.

A large loss in on-heating ductility was observed at ~2100 F that was determined to be directly related to boron content. Constitutional liquation of borides could cause partial grain boundary liquation, which would result in a fracture mode change and reduction in strength above about 2100 F; however, constitutional liquation was not directly observed. This absence in ductility and loss in strength between ~2000 F and the zero strength temperature of about 2300 F leaves this region of the HAZ very susceptible to hot cracking.

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Introduction

A-286, an age-hardenable austenitic stainless steel, is subject to weld hot cracking because of its susceptibility to both fusion zone cracking and heat-affected zone (HAZ) microfissuring. The HAZ cracks, intergranular in nature and often only several grain diameters long, are usually undetectable by nondestructive testing techniques.

Rather extensive studies have been conducted on A-286 to determine the weld cracking mechanisms. Vagi and Martin (Ref. 1), through phase extraction techniques on samples heated to 2450 F (1343 C), associated the HAZ microfissuring with formation of the Fe_2Ti Laves phase. They proposed that grain boundary liquation by the Fe_2Ti-Fe eutectic (2390 F, 1310 C) allows the grains to be separated by stresses of thermal origin. Blum and Witt (Ref. 2) agreed with Vagi and Martin, but also proposed that grain boundary precipitation of a continuous film of TiC and precipitation hardening by γ', $Ni_3(Ti, Al)$, inhibits deformation during cooling, which causes base metal cracks to propagate to relieve the weld stresses. Radavich (Ref. 3) also concluded from elevated temperature phase stability studies that TiC plays a role in weld cracking. He proposed that films of grain boundary carbides thicken and either force the grains apart or leave them enbrittled and susceptible to cracking under welding stresses.

The purpose of this study was to determine the effect of alloy chemistry on HAZ microfissuring. If the relationship between the chemistry and the cracking mechanisms is understood, it may be possible to make chemistry modifications that would reduce HAZ cracking susceptibility without sacrificing mechanical properties. Support for this suspected relationship was provided by a previous study (Ref. 4) which demonstrated that slight chemistry modifications significantly reduced the susceptibility of A-286 stainless steel to fusion cracking.

In the present study, the effect of chemistry on the HAZ cracking mechanisms was studied by means of spot varestraint and hot ducility testing. In prior studies, the hot ductility testing technique has proven most valuable in studying the hot-cracking mechanisms that occur during welding (Refs. 9-12). In the present study, this technique was most beneficial in understanding the role that different elements play in the hot-cracking mechanisms. The spot varestraint tests provided a direct comparison between HAZ hot cracking and hot ducility response.

Table 1 — A-286 Experimental Heats

Heat no.	Chemical composition, wt.%, balance Fe [a]											
	C	Mn	P	S	Si	Cr	Ni	Al	Mo	V	Ti	B
1	0.063	1.50	0.004	0.005	0.58	14.45	25.40	0.22	1.27	0.26	2.25	0.0073
2	**0.031**	1.51	0.004	0.006	0.63	14.50	25.70	0.24	1.27	0.26	2.22	0.0059
3	**0.025**	1.47	0.004	0.005	0.61	14.50	25.50	0.25	1.26	0.25	2.20	**0.0009**
4	**0.028**	**0.13**	0.006	0.005	**0.25**	14.77	26.01	0.23	1.25	0.26	2.18	0.0063
5	**0.027**	**0.11**	0.004	0.005	**0.16**	14.70	24.70	0.10	1.25	0.25	2.18	**0.0014**
6	**0.028**	1.54	0.004	0.005	0.63	14.25	**34.70**	0.25	1.24	0.25	2.15	0.0064
7	**0.025**	1.51	0.004	0.005	0.60	14.75	25.45	**0.66**	1.27	0.23	**1.60**	**0.0012**
8	**0.025**	1.39	0.010	0.008	0.58	15.20	23.50	0.24	1.25	0.25	**2.63**	0.006
9	0.068	1.32	0.018	0.003	0.63	14.90	24.93	0.19	1.25	0.21	2.15	0.004

(a) Bold face numbers indicate major chemistry modification

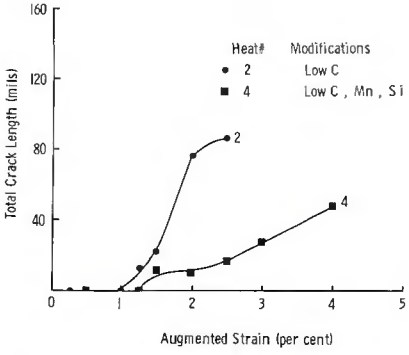


Fig. 1 — Typical spot varestraint data

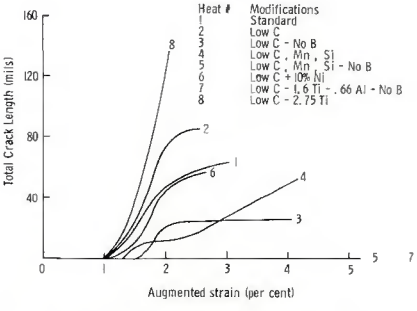


Fig. 2 — Spot varestraint results showing relative HAZ hot cracking susceptibility

Material

Eight 50 lb VIM (vacuum induction melt) experimental heats and one commercial heat were used to study the effect of alloying elements on weldability (see Table 1). Before testing, the material was solution treated for one hour at 1800 F and water quenched; this process produced a grain size of about ASTM 6.

Spot Varestraint Testing

Procedures

Spot varestraint testing was conducted on heats 1-8 of Table 1 using the testing procedure described in Refs. 5, 6, and 7. In summary, the test utilizes a GTA gun to melt a spot in the center of a 2.54 × 15.24 × 0.64 cm sample. After a predetermined time, the torch is extinguished and a ram beneath the sample is actuated. The ram, holding an interchangable radius die block, rapidly deforms the sample into the shape of the die. The augmented tangential strain is $\varepsilon \simeq t/2R$, where t = specimen thickness and R = die block radius. For these tests a delay timer was set such that the ram struck the sample 140 msec after the welding arc was extinguished. A welding time of 8 sec at 130 A and 8 V was used with argon shielding.

The total length of weld cracking in only the HAZ was measured under 30 or 60X magnification. This crack length total was then plotted for different strain values.

Results

The curves for heats 2 and 4 (Fig. 1), which are typical of the varestraint results obtained, are similar in shape to the curves exhibited by other alloys tested in this manner (Ref. 6). Figure 2 compares the relative hot cracking susceptibility of all the experimental heats. (For clarity, the data points have been excluded. However, all of the data fit the curves to the same degree as shown in Fig. 1.) As shown in Fig. 2, the increase in the Ti content from 2.22 to 2.63% in standard A-286 is very detrimental to HAZ hot cracking. Also, the decrease in carbon from 0.063 to 0.031% in standard A-286 has some detrimental effect. The decrease in Mn and Si (heat 4), the decrease in B from 0.0059 to 0.0009% (heat 3), and the direct substitution of 0.4% Al for Ti all had beneficial effects on the reduction of HAZ microfissuring. However, the 10% increase in the Ni content had very little if any significant effect. Little if any cracking was observed in heat 5 (low C, Mn, Si, and no B) and heat 7 (low C, 1.6 Ti, 0.66 Al, and no B) even after 8% augmented strain. With a shorter ram delay time some cracking would be expected, but the relative hot cracking susceptibility should remain unchanged.

Hot Ductility Testing

This testing technique, developed at Rensselaer Polytechnic Institute and described in Ref. 10, reproduces in a test specimen the time-temperature cycle experienced at any given point in the weld HAZ. The specimen, which is resistance heated, can also be tested in tension to failure at any time during the thermal cycle. In this study, specimens were pulled "on-heating" at various peak temperatures to establish the nil ductility temperature, NDT (the temperature at which the ductility measured by the reduction in area at the fracture surface is only 2-3%), and the zero strength temperature, ZST (the temperature at which the strength approaches zero). Specimens were also tested during the cooling portion of the thermal cycle, "on-cooling," after experiencing heating to a given peak temperature, in most cases the ZST.

Much information has been gained in the past fifteen years concerning the use of hot ductility testing to predict the susceptibility of a material to HAZ hot cracking during welding. However, since this technique has not been standardized, the optimum testing procedures are still questionable. The correlation between the HAZ cracking experienced in actual welds and hot ductility testing is better when testing on-cooling from the ZST than from the ZDT (Refs. 8, 9). Even when the ZST is used, one often has to decide whether the recovery of strength and ductility are sufficient to prevent HAZ cracking (Ref. 8).

Procedures

Hot ductility testing was conducted using a Gleeble Model 510 testing machine with a chromel-alumel thermocouple accurate to 0.3% millivolt reading. Both on-heating and on-cooling tests were conducted on ¼-in. diam samples of all nine heats, over the thermal cycle shown in Fig. 3. The on-cooling experiments on the commercial heat were conducted from three peak temperatures — 2125, 2200, and 2375 F — and from slightly below the ZST on the experimental heats.

On-Heating Results

The on-heating ductility of the commercial alloy dropped drastically at 2100 F (1149 C), as emphasized by the large number of data points in Fig. 4. The ill-defined NDT is approximately 2150 F.* The standard experimental chemistry, heat 1, exhibited this same ductility behavior (see Fig. 5a). In addition, the strength of heat 1 decreased continuously from 1800 F to the ZST of about 2275 F (Fig. 5b), in the same manner as heat 9 (not shown). As shown in Fig. 5b, the reduction in carbon from 0.063 to 0.031 in heat 2 may have shifted the ductility drop about 50 F higher in temperature (compared at 30% RA), but there is no significant difference in either the ductility or strength (Fig. 5c) of the two carbon levels.

As shown in Fig. 6a, the increase of 9% Ni, heat 6, increased the ductility drop approximately 25 F, but had no significant effect on either the on-heating strength or ductility. The increase of 0.4% Ti, heat 8, reduced the 2100 F ductility drop almost 100 F (Fig. 6c). Also, the ZST was reduced by 100 F to about 2200 F.

The reduction In B to 0.0009% (heat 3) had a large effect on both the on-heating strength and ductility. The 2100 F ductility drop was absent (Fig. 7a). Instead, relatively good strength and ducility were maintained to within 50-75 F of the ZST, which was increased about 40 F to 2330 F (Fig. 7b). As shown in Figs. 7c through 7f, the on-heating behavior of heat 7 (low C, B, +0.4 Al, and −0.6 Ti) and heat 5 (low C, B, Si, and Mn), were very similar to the low B heat 3. Again, the drop in ductility around 2100 F was absent, and both heats experienced relatively good strength and ductility to the ZST of about 2325 F.

An on-heating anomaly was observed in both the strength and ductility of heat 4 (low C, Mn, and Si). The large drop in ductility again occurred at 2100 F, the same as in the standard low C heat 2. However, between 2150 and 2200 F the ductility started to recover from 5 to 30%, but then dropped to the NDT of 2275 F, as shown in Fig. 8a. A corresponding minimum in strength was also observed (Fig. 8b). A second set of tests confirmed the minimum experienced in the strength and ductility of this heat.

The NDT temperature will be used instead of the ZDT (Zero ductility temperature) since the ZDT is not well defined in these alloys

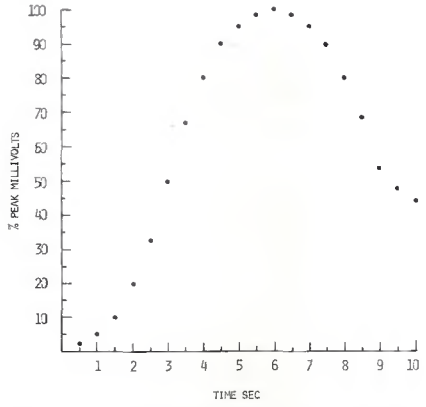


Fig. 3 — Heating cycle used in hot ductility tests

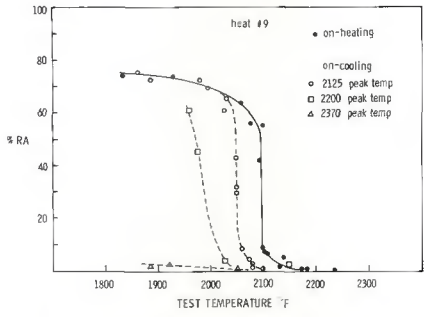


Fig. 4 — Hot ductility results for commercial A-286 — note abrupt ductility loss at 2100 F

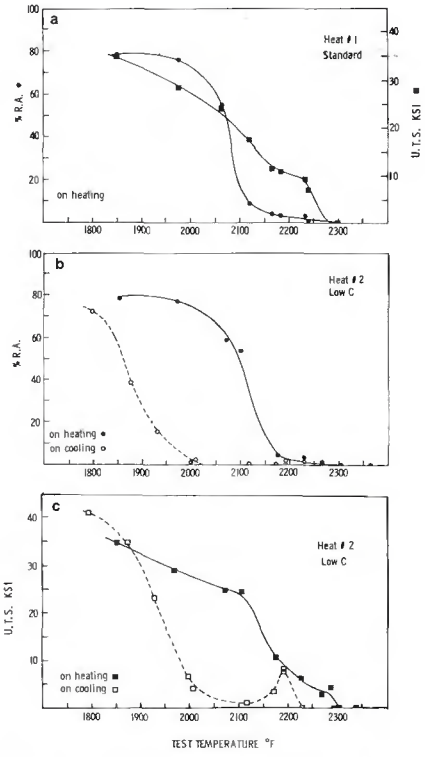


Fig. 5 — Hot ductility data for two carbon levels

The on-cooling ductility of heat 9 (Fig. 4) was recovered at lower temperatures as the peak temperature was increased above the large on-heating ductility drop. It became increasingly apparent that this heat is susceptible to hot cracking. However, for a peak temperature of 2375 F, which is approximately 75 F above the ZST, little ducility was recovered even as low as 1875 F. These results emphasized the importance of peak temperature on the on-cooling results. Therefore, all the on-cooling testing of the experimental heats was conducted from the ZST.

On-Cooling Results

Heat 2 (low C) recovered little strength or ductility as much as 300 F below the ZST (Figs. 5a and 5b). Heat 8 (low C, +0.4% Ti) also recovered little ductility until 250 F below the ZST (Fig. 6b). However, a significant amount of strength was recovered 100 F below the ZST (Fig. 6c). Heat 4 (low C, Mn, and Si) recovered little ductility until about 450 F below the ZST (Fig. 8a); the recovery was similar to that of heat 2. The recovery of strength was also extremely poor, but somewhat better than heat 2.

The on-cooling behavior of heats 3, 5, and 7 (Fig. 7) was similar. These heats recovered significant strength and ductility within 100-150 F below the ZST. Of the three heats, heat 5 recovered both strength and ductility most rapidly when compared, for example, at 2200 F.

A summary of the above hot ductility data is shown in Table 2.

Discussion

Since stresses can be relieved both plastically and elastically, it is necessary to examine the on-cooling recovery of both strength and ductility when predicting hot cracking susceptibility from hot ductility results. The heats studied can easily be separated into two groups based upon their hot ductility response. Heats 1, 2, 4, 6, and 8 all exhibited a large loss in on-heating ductility around 2100F and showed very poor recovery of both

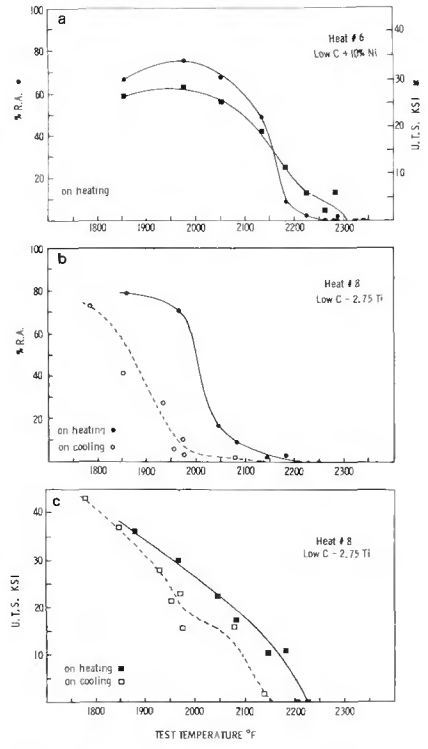


Fig. 6 — Hot ductility results

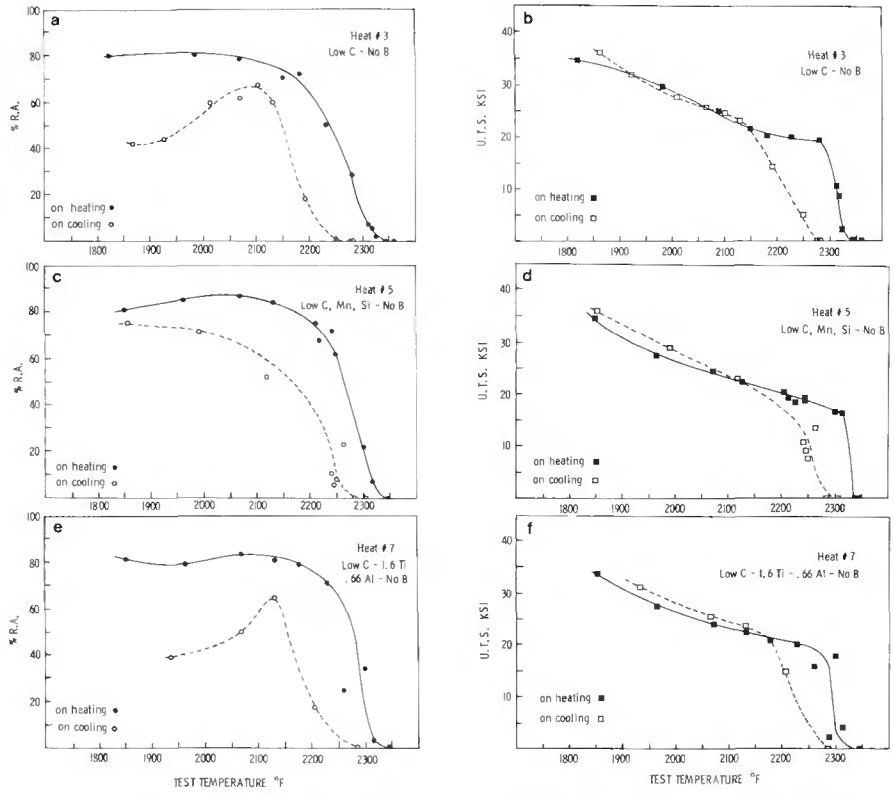


Fig. 7 — Hot ductility results for low B heats — note absence of ductility loss at 2100 F

Table 2 — Summary of Hot Ductility Data from Figs. 5-8

Heat	Alloy modification	On-heating temp., F (a) 30% RA	10 ksi UTS	ZST	On-cooling temp. F 30% RA	10 ksi UTS
1	Standard	2085	2210	2275	—	—
2	low C	2120	2180	2290	1895	1980
3	low C, no B	2275	2315	2330	2170	2220
4	low C, Mn, Si	2120	2265	2340	1880	2045
5	low C, Mn, Si, B	2285	2325	2340	2225	2250
6	low C + 10% Ni	2160	2185	2300	—	—
7	low C, 1.6 Ti, 0.66 Al, no B	2285	2290	2325	2180	2220
8	low C, 2.63 Ti	2020	2180	2200	1910	2090

(a) Temperatures are only approximate, as can be seen by examining the actual data in Figs. 5-8.

strength and ductility on-cooling from the ZST. Therefore, it is predicted that these heats are all very susceptible to HAZ hot cracking, as verified by the spot varestraint tests. Heats 3, 5, and 7 all recovered significant on-cooling strength and ductility within 100-150 F of the ZST. Therefore, these heats are much less susceptible to HAZ cracking than the first group. Based upon the hot ductility data, it becomes increasingly difficult to confidently predict which heat within each group is less susceptible to HAZ cracking. However, the spot varestraint tests show the relative hot cracking susceptibility of all the heats (Fig. 2) and clearly indicate the beneficial or detrimental effect of all the alloy modifications.

2100 F Ductility Loss

The large loss in on-heating ductility at 2100 F corresponds to a change in primary fracture mode from transgranular ductile behavior to intergranular fracture. The complete intergranular failure of heat 9 at 2150 F is shown in Fig. 9a. However heats 3, 5, and 7, which do not contain intentional additions of boron, exhibited good ductility and a large amount of transgranular rupture within 50 F of the ZST (Fig. 9b). The hot ductility tests clearly indicate that the boron content rather than grain growth (Ref. 2) is responsible for both the ductility drop and change in fracture mode at approximately 2100 F.

It seems very possible that the ductility drop may be explained by constitutional liquation of grain boundary borides under the rapid non-equilibrium heating rates. Constitutional liquation results from non-equilibrium heating conditions, which cause rapid decomposition of unstable intermetallic compounds. This decomposition highly enriches the region adjacent to the dissolving particles in solute. The equilibrium phase for this high solute composition is a liquid that surrounds the particle. The liquid region will remain until the particle completely dissolves and the solute diffuses from the remaining liquid to a composition less than the equilibrium solidus composition.

The constitutional liquation of carbides (Refs. 10-13, 15) and sulfides (Ref. 14) has been proposed as the mechanism of HAZ cracking in several different alloy types. Owczarski et al (Ref. 12) have attributed the constitutional liquation of M_3B_2 to be largely responsible for HAZ cracking in Udimet 700. In the present study, borides were not directly observed; however, the boron content of A-286 is only about 20% of that of Udimet 700. If constitutional liquation does occur around 2100 F, the liquid phase should eventually disappear after sufficient boron diffusion results. This was found to be the case in Ni 18 maraging steel in which the disappearance of evidence of constitutional liquation was complete after 6 sec at 2450 F (Ref. 14).

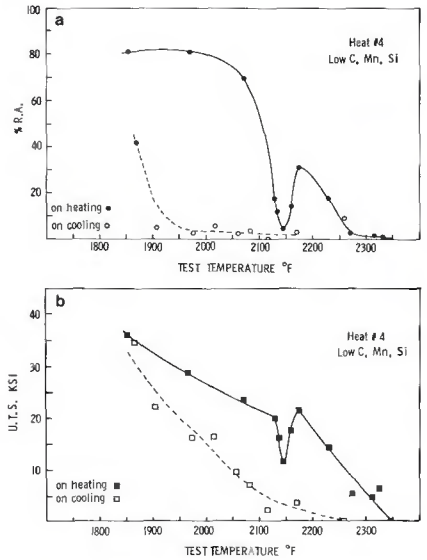
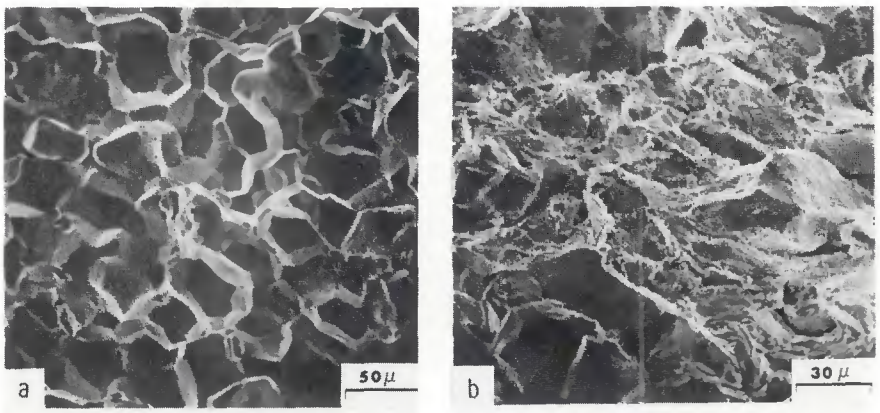


Fig. 8 — Hot ductility results — note on-heating anomaly



a. Heat 9, 2150°F, RA = 2.5% - total intergranular fracture

b. Heat 5, 2315°F, RA = 7% - exhibits partial ductile transgranular fracture

Fig. 9 — Fracture surfaces of hot ductility samples

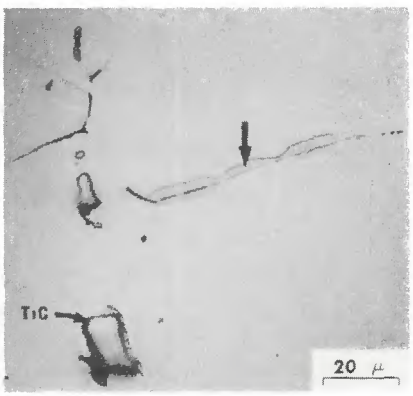


Fig. 10 — Area of apparent grain boundary liquation in heat 9 heated to a peak temperature of 2200 F

Hot ductility specimens were heated as before and held at a peak temperature of 2200 F for times of 0 to 1200 sec. In no case was any ductility or strength recovered. The absence of ductility recovery suggests that either sufficient boron diffusion had not occurred for total disappearance of liquation, or that a liquid phase was in equilibrium at that temperature. If the latter is true, then, theoretically, constitutional liquation is not involved.

In either case, grain boundary liquation is promoted at approximately 2100 F by the presence of boron. Boron does form a eutectic with Fe at 2100 F and has very limited solid solubility in Fe (Ref. 16). Some areas of what appeared to be grain boundary liquation were observed in the samples held at 2200 F (Fig. 10); however, electron microprobe studies failed to identify any composition change. It appears that at approximately 2100 F enough grain boundary liquation occurs to promote complete intergranular failure and a large loss in strength. However, enough grain boundary area remains free of liquid to maintain some strength to the ZST of about 2275 F.

ZST

The loss of strength and ductility at around 2300 F is associated with the onset of massive grain boundary liquation. Scanning electron microscopy (SEM) examination of hot ductility specimens of heat 9 showed that the melt phase nucleates largely around decomposing Mo enriched TiC or Ti(C,N). This phase started to form around 2275-2300 F, the ZST. At about 2375 F all the carbides had dissolved, leaving the grains surrounded with the melt phase (see Fig. 11).

Metallographic examination of spot varestraint specimens showed the association of grain boundary liquation by this phase with HAZ cracking (Fig. 12a-b). The SEM energy dispersive x-ray spectrometer (Fig. 12c-d) showed this phase to be enriched, with respect to the matrix, in Si, Ti, Ni and Mo. X-ray techniques confirmed that this phase is the Fe_2Ti type Laves phase (Ref. 4), but this qualitative analysis shows this phase to be more complex than the simple Fe_2Ti originally reported (Ref. 1). The Fe_2Ti-Fe eutectic temperature is about 2390 F. However, because of the complex nature of the Laves phase, compared to pure Fe_2Ti, slight alloy modifications can alter the phase equilibria, properties, and quantity of phase formed. Examples of this type of alteration are shown indirectly by the changes in high temperature strength, ductility, and especially the ZST of these alloys.

The increase in Ti in heat 8 greatly increased the amount of melt phase formed in the HAZ of both spot varestraint and electron beam welds. The reduction in the ZST by 100 F that accompanied the addition of 0.4% Ti also indicates that a significant amount of the melt phase forms at a lower temperature. These two factors can explain the increased HAZ cracking observed in spot varestraint tests.

Since the Laves phase is enriched in Si, it is not surprising that modifications in Si affect cracking. The 50 F increase in the ZST with low Si and Mn (heat 4) indicates that the melting point is lowered by Si. Boron also appears to lower the melting point of the melt phase since the reduction of B (heat 3) increased the ZST by 50 F. The effect of B and Si may be expected since both elements form low melting eutectics with both Fe and Ni.

Hot Cracking Mechanism

The loss in on-heating ductility and reduction of strength between approximately 2100 F and the ZST is due to some degree of grain boundary liquation promoted by B, possibly constitutional liquation of borides. At about 2300 F the Laves phase formation results in complete grain boundary liquation at the ZST of approximately 2300 F. The presence of the grain boundary liquation in the region of the HAZ above 2100 F leaves this region easily susceptible to microfissuring under the weld stresses. On-cooling experiments from above the ZST show that the Laves phase severely embrittles the grain boundaries at temperatures as low as 1875 F. The strength is also lowered about 50% at 1875 F on-cooling from 2370 F.

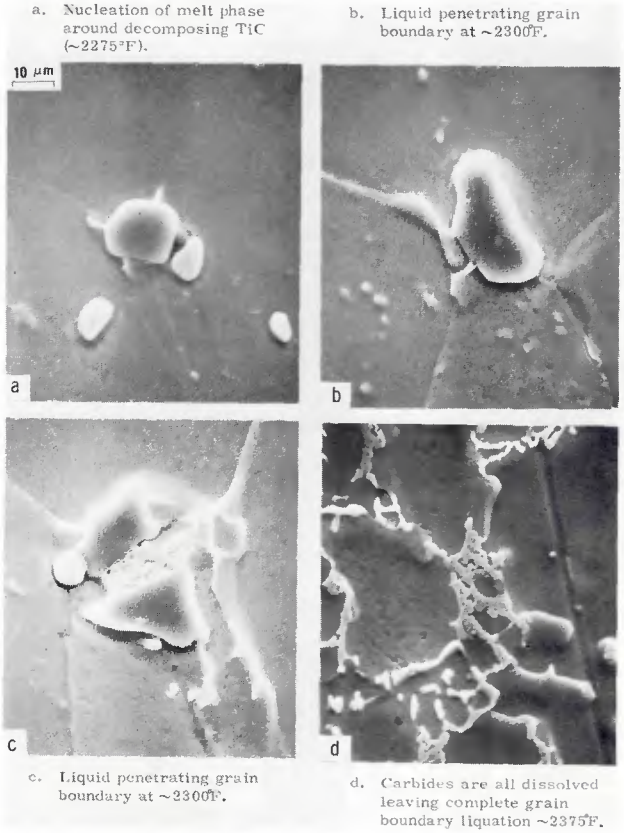


Fig. 11 — Hot ductility sample heat 9 — Laves phase formation

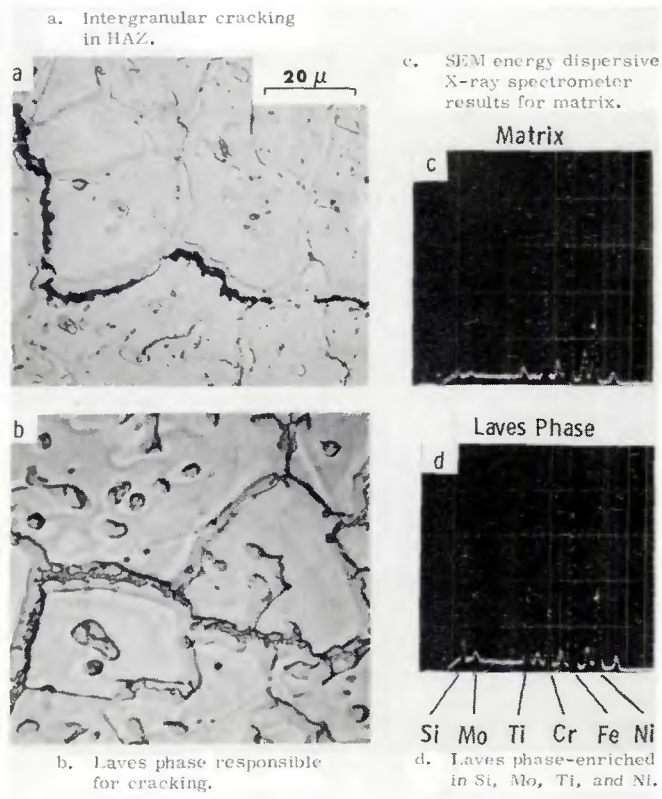


Fig. 12 — Weld HAZ of spot varestraint sample

Therefore, it seems very likely that HAZ cracking can also occur in the solid state. Cracks that form while liquid is still present could easily propagate further in the solid state along the embrittled grain boundaries as the solidification stresses increase. As the M_2Ti-Fe eutectic temperature decreases, the heavily liquated region of the HAZ, which is highly susceptible to cracking, becomes wider. Thus the amount of HAZ cracking would be expected to increase, as was the case for heat 8. However, the properties of that region of the HAZ which experienced peak temperatures between about 2100 F and the ZST had the largest effect on HAZ hot cracking. The heats which had relatively good strength and ductility in this region (heats 3, 5, and 7) exhibited little HAZ cracking during spot varestraint testing, even in the regions of heavy grain boundary liquation. This reduction in cracking indicates that sufficient stresses can be relaxed in this region of the HAZ to largely prevent cracking over the entire HAZ region. However, in highly confined welds, some cracking in the liquated region must be expected even in heat 5.

On-Heating Anomaly

The on-heating anomaly of heat 4 is not thoroughly understood. The rapid constitutional liquation of borides could have caused the drop at 2100 F as before. In this case, the boron would have rapidly diffused away, allowing the strength and ductility to recover. Then, at higher temperatures, the Laves phase would have started to form, with resultant losses in strength and ductility. However, samples of heat 9 held at 2200 F for 0 to 1200 sec showed no recovery in strength or ductility with time. At 2280 F the Laves phase was very visible. It is possible that a small amount of Laves phase is formed at 2200 F when Si is present and that it prevents recovery from the constitutional liquation.

It was shown that the Laves phase is enriched in Si and also that the reduction of Si increased the ZST by about 50 F. Boron is also probably associated with the Laves phase, since the reduction of boron increased the ZST and also greatly reduced fusion zone cracking, which is also due to Laves phase formation (Ref. 4). Therefore, for standard B and Si (heat 2) the constitutional liquation and Laves phase formation may occur at sufficiently close temperatures that no recovery in strength and ductility occurs above 2100 F, as in the case of heat 4. However, more experiemntal data is necessary to confirm this hypothesis.

Summary

Hot ductility testing proved very valuable in understanding the hot cracking mechanisms involved in HAZ cracking of A-286. Based upon their hot ductility behavior, the alloys studied can be divided into two groups with vastly different HAZ cracking susceptibility. However, within each group it becomes very difficult or impossible to order the alloys on the basis of their hot cracking susceptibility.

The spot varestraint tests clearly show the relative hot cracking susceptibility of all the heats and clearly indicated the effect of slight alloy modifications on hot cracking susceptibility.

Conclusions that could be drawn from the hot ductility tests and subsequent microstructural examination of test specimens are:

1. A sharp loss in on-heating ductility occurs at ~2100 F in standard A-286. This loss is clearly associated with the presence of B and is possibly due to the constitutional liquation of borides resulting possibly in the formation of Fe_2B-Fe eutectic at about 2100 F.

2. The ZST is due to the Laves phase formation which results in subsolidus grain boundary liquation.

3. The Laves phase nucleates largely around decomposing TiC.

4. The Laves phase is enriched in Ni, Si, Ti and Mo with respect to the matrix.

5. The complex nature of the Laves phase allows one to change the amount and properties of the melt phase formed through slight alloy modifications.

6. Reduction in especially B and Si greatly improve hot ductility behavior.

Spot varestraint tests showed that:

1. Cracks formed not only in the heavily liquated region of the HAZ, but also in regions further away from the fusion zone in which hot ductility tests showed little ductility and reduced strength.

2. Increasing Ti from 2.22 to 2.63 wt.% greatly increased the hot cracking tendency.

3. The reduction of B from 0.006 to 0.001% greatly reduced the hot cracking tendency.

4. The reduction of Mn and Si significantly reduced hot cracking. Si is probably responsible for the beneficial effects since it is contained in the Laves phase.

5. The direct subsitution of about 0.4% Al for Ti reduced the cracking susceptibility in the absence of B.

6. Reduction of carbon from 0.063 to 0.031% increased the hot cracking susceptibility, which indicates that the formation of TiC does not play as important a roll in cracking as earlier reported.

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