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The Influence of Yttrium Microadditions on Titanium Weld Metal Cracking Susceptibility and Grain Morphology

Controlled additions of yttrium contributed to changes in the fluid flow and heat extraction in the weld pool

BY M. C. NORDIN, G. R. EDWARDS AND D. L. OLSON

ABSTRACT. Yttrium additions were made to Ti-6Al-2Nb-1Ta-1Mo weld metal in attempts to refine the grain structure and reduce the cracking susceptibility. Yttrium contents in the fusion zone as low as 0.01 weight-percent were shown to eliminate cracking. This reduction in cracking susceptibility was related to a refinement of the fusion zone grain size microstructure. Yttrium additions were observed to increase the turbulence of the fluid flow in the weld pool. Decanting of the weld pool showed that this turbulent flow drastically altered the bead morphology in a way which correlated with the observed change in grain morphology.

Introduction

Titanium weld metal microstructure and properties are dependent on numerous parameters, including the welding process used, the heat input, and the cooling rate. In titanium weld metal, both grain size and shape can be important microstructural features. Poor fusion zone ductility has been attributed to large beta grains in GTA weldments of Ti-6Al-4V, Ti-6Al-2Sn-4Zr-6Mo, and Ti-6Al-6V-2Sn (Ref. 1). Large beta grain size can result in beta grain boundaries which completely traverse the fusion zone, thus providing a preferential crack propagation path. Large grains often result from epitaxial grain growth, both in titanium weld metal and in the weld metal of other alloy systems (Refs. 1–8). The partially melted grains at the solid-liquid interface act as ideal substrates upon which growth of the solid phase into the molten pool can occur. Grains then grow epitaxially from the FZ/HAZ interface and may continue to grow completely across the fusion zone. During multipass welding of titanium alloys, these long columnar grains continue to grow from bead to bead, resulting in grains that traverse the complete weld. A major objective in single- and multipass titanium welding, therefore, is to prevent epitaxial grain growth and refine the fusion zone grain size.

The refinement of the fusion zone grain structure can be achieved by many methods (Ref. 6). Nucleating agents are widely used to refine the grain size in castings. However, very limited data is available on the use of inoculants to refine weld metal.

Attempts have been made to inhibit epitaxial grain growth and/or refine the prior beta grain size in titanium weldments using high-temperature inoculants such as yttrium. Simpson investigated the effects of yttrium inoculation on weld pool structure and the resultant mechanical properties (Ref. 9). Yttrium was added in hopes that yttria particles would form and reduce the grain size by acting as heterogeneous nucleation sites. Yttrium was found to successfully reduce the fusion zone grain size and thereby improve thin sheet weldability. However, epitaxial growth into the weld metal from the weld interface was not inhibited by yttrium additions, and grain refinement was obtained only at the center of the weld. Simpson observed high concentrations of yttria particles at the grain boundaries, and concluded that the particles were pushed ahead of the advancing solid-liquid interface. The resultant drag effect was thought to increase as solidification continued and the yttria particle density increased, providing more heterogeneous nucleation sites for the last metal to solidify.

KEY WORDS

Yttrium-Titanium
Yttrium Microaddition
Titanium Weld Metal
Weld Metal Cracking
Grain Nucleation
Yttrium Doping
Weld Grain Refinem't
Weld Grain Growth
Weld Pool Turbulence
Grain Morphology

Other more recent attempts (Refs. 5, 6) also have been made to inhibit epitaxial grain growth and/or refine the prior beta grain size in titanium welds by adding yttrium to the weld metal. These studies, which employed the multiple-pass GTA welding process to weld Ti-6Al-4V, again showed that small amounts of yttrium did not inhibit epitaxial grain growth, but refined the grains in the center of the fusion zone. In the study of Misra, *et al.* (Ref. 6), a uniform distribution of yttria particles was observed throughout the fusion zone. This was in contrast to the observation of Simpson (Ref. 9), who attributed grain refinement to the buildup of yttria particles at the grain boundaries during solidification. The mechanism responsible for the grain refinement caused by yttrium additions is not immediately obvious. Conceivably, yttrium added to the weld pool could either provide more sites for heterogeneous nucleation of grains, or it could retard the grain growth kinetics (Ref. 6).

The growth rate of the solidification front, R, can be determined from knowledge of the torch velocity, ν, and the angle, Θ, between the columnar grains and the direction of welding, from:

$$R = \nu \cos \Theta \qquad (1)$$

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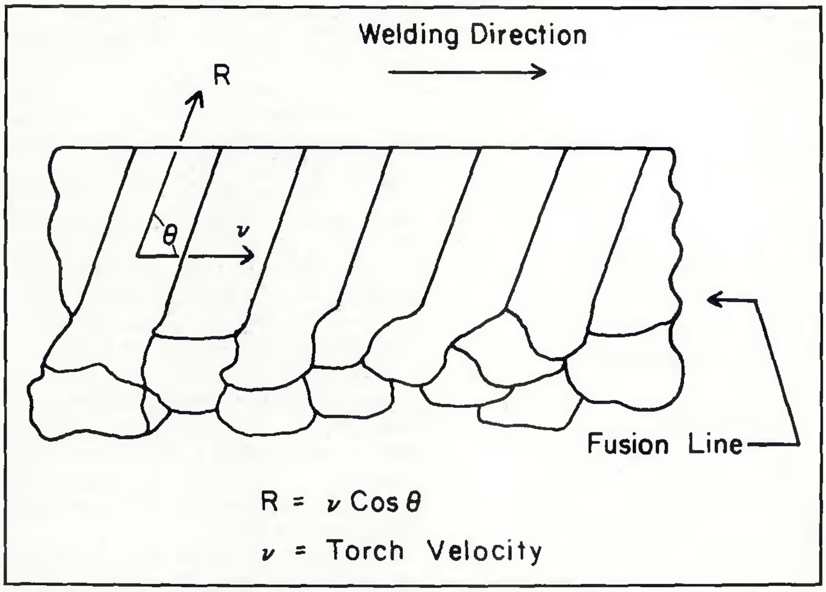

Fig. 1 — Schematic diagram of a longitudinal section through the fusion zone showing the growth rate of columnar grains, R

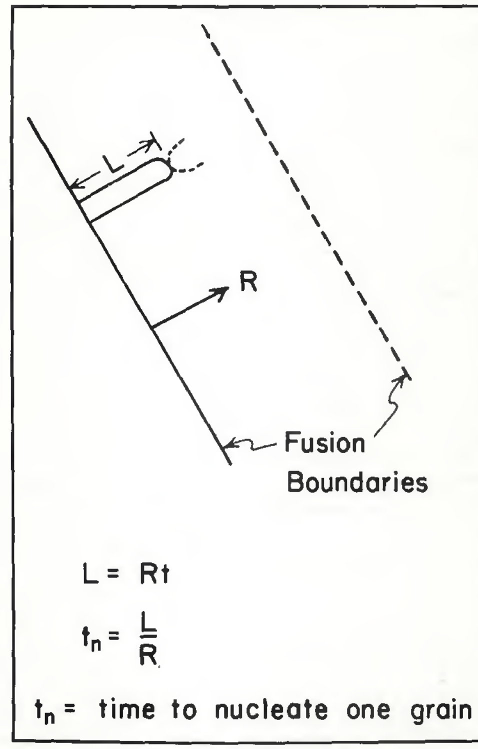

Fig. 2 — Schematic diagram illustrating the method used to determine the time to nucleate the first grain, t_n, in the weld pool

The schematic representation of R, ν and Θ are shown in Fig. 1. Misra, *et al.* (Ref. 6), found that as the yttrium content increased in the fusion zone, the angle, Θ, decreased, and thus the growth rate increased. An increase in growth rate should promote long columnar grains in the fusion zone in the absence of an increase in nucleation rate.

The growth time prior to nucleation of a new grain, t_n, was determined from knowledge of the growth rate and the distance, L, from the fusion line to the first equiaxed grain, as shown in Fig. 2. The time, t_n, is also the incubation time prior to nucleation of the new grain which terminates growth; hence the nucleation time. Linear growth rates were assumed, so that:

$$L = R\, t_n \qquad (2)$$

Misra, *et al.* (Ref. 6), found that the time to nucleate a new grain was reduced by an order of magnitude with an addition of only 0.1 at.-% yttrium. It became apparent, therefore, that even though growth rate was increased, the nucleation rate was the dominant effect, and grain refinement resulted. Fundamental concepts of nucleation and growth were then applied to explain the grain refinement mechanism. Integration of expressions for both the nucleation rate and growth rate provided two competing times:

1) t_n = time to nucleate one grain
2) t_g = time for one grain to grow across the weld bead

These two times are expressed in the following forms:

$$t_n = C_3 \exp\left[\frac{C_2}{\Delta T^2}\right] \qquad (3)$$

$$t_g = \frac{x_0}{C_1\, \Delta T} \qquad (4)$$

where x_0 is the bead depth, ΔT is the degree of undercooling, and C_2 and C_3 are constants. The ΔT terms in Equations 3 and 4 may not be identical. The undercooling associated with growth may be affected by capillarity and solute partitioning. The small difference between the ΔT terms does not affect the conceptual thinking, however, and Equations 3 and 4 can be schematically compared as shown in Fig. 3.

Two values of ΔT must be considered to determine whether nucleation will occur before a columnar grain grows across the fusion zone. One, ΔT_c, is the critical initial undercooling required for rapid nucleation. The second, ΔT_w, is the undercooling achieved during the welding process. If $\Delta T_c < \Delta T_w$, then new grains will nucleate before they grow across the fusion zone, *i.e.*, $t_n < t_g$. With yttrium additions, $t_n < t_g$ and the coarse columnar fusion zone grain structure is broken up.

It was proposed that yttrium addition shifted the nucleation curve in Fig. 3 to the left, resulting in a $\Delta T_c < \Delta T_w$ and grain refinement. This effect of yttrium was fundamentally interpreted to be related to a change in either the liquid-solid interfacial energy, σ, or the contact angle, Θ. Both of these parameters are physical properties contained in constant C_2. This model would suggest that yttrium-bearing particles act as nuclei, reducing ΔT_c and providing sites for heterogeneous nucleation of new grains. However, grain refinement was obtained only at

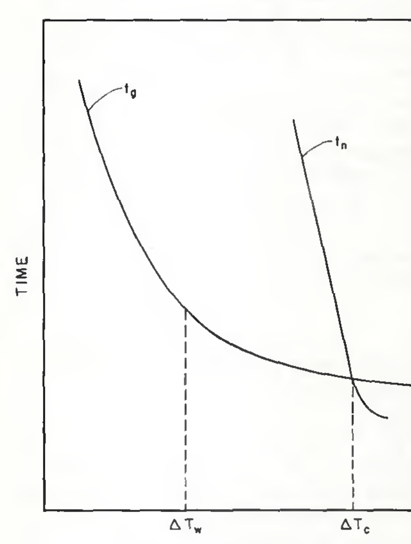

Fig. 3 — The nucleation and growth times versus undercooling during weld pool solidification, where t_g is the time to grow across the bead and t_n is the time to nucleate one grain

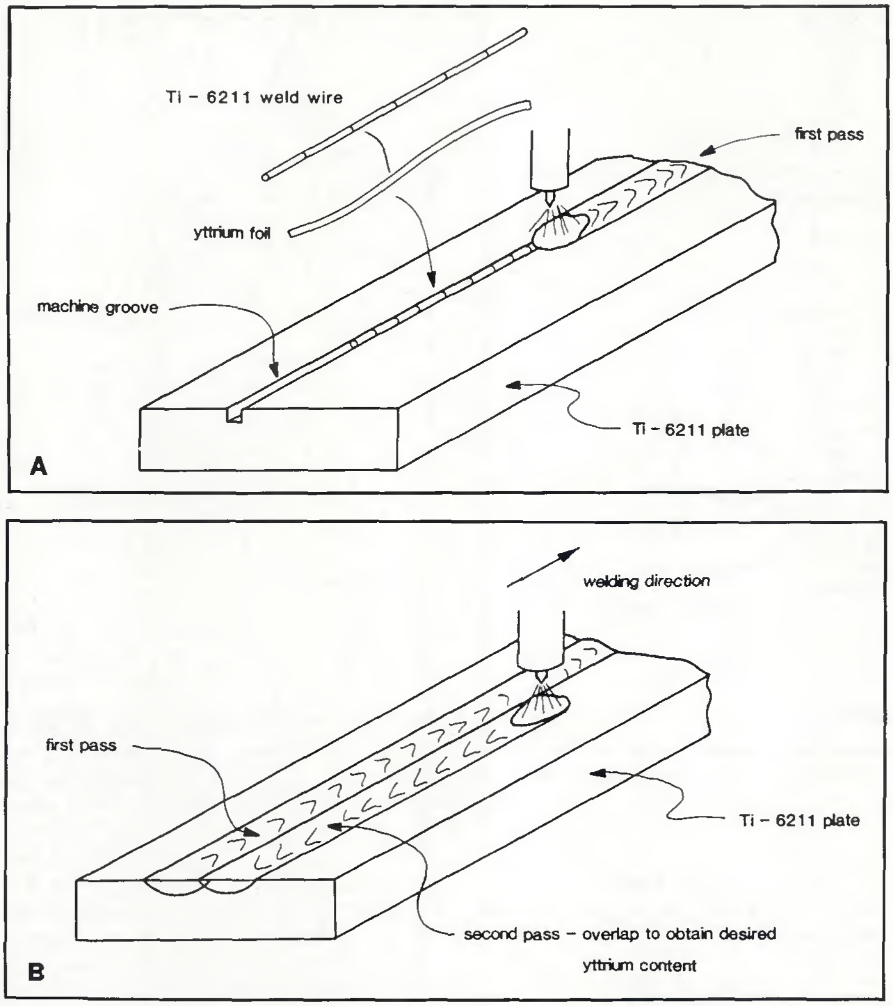

Fig. 4—A—Procedure for yttrium doping of Ti-6211 weld metal using the GTA welding process; B—procedure for yttrium dilution into fusion zone to obtain desired yttrium content

the center of the weld, not near the weld interface, even though yttrium-bearing particles were seen to be uniformly distributed throughout the fusion zone. If yttrium only acted to provide heterogeneous nucleation sites, more uniform grain refinement in the fusion zone would be expected. Thus, the exact mechanism by which yttrium causes grain refinement remains somewhat in question. The present study seeks to clarify the role which these yttrium microadditions play in the solidification of titanium weld metal.

Experimental Procedure and Analysis

The material used for this study was 13-mm (0.51-in.) thick Ti-6211 beta-processed plate. The chemical composition of the base plate is given in Table 1. GTA autogenous welds were made on 13- × 30- × 280-mm (0.51- × 1.2- × 11-in.) specimens. Weld metal yttrium contents from zero to 0.24 wt-% were studied. To obtain a controlled level of weld metal yttrium, a two-pass dilution technique was used, as illustrated in Figs. 4A and 4B. First, 1- × 1-mm (0.04- × 0.04-in.) grooves were machined into each plate. Next, a 0.13- × 0.6-mm (0.005- × 0.02-in.) wide yttrium foil was cleaned, then positioned within the groove for the entire length of the plate. A 1-mm-diameter Ti-6211 wire was then press fit into the groove to hold the yttrium foil in place. The foil and weld wire were melted with one pass of the GTA torch, and the sample was allowed to cool to room temperature. A second overlapping pass was then made (see Fig. 4B) to provide weld metal with the desired yttrium content. An electron beam microprobe was used to determine the actual weld metal yttrium content.

Direct current with electrode negative polarity was used for welding. Two percent thoriated tungsten electrodes, 3.2 mm (⅛ in.) in diameter and ground to a 45-deg conical tip, were used, and the electrode gap was set at a distance of 1.5 times the electrode diameter. Commercial grade argon with 30-cfh (14-L/min) flow rate was used as a cover gas for all weldments. The GTA welding parameters were held constant at 14 V, 180 A and a travel speed of 4 ipm (1.7 mm/s). The heat input was 1.5 MJ/m (38 kJ/in.). Arc stability measurements were made using a Tektronix 4103N oscilloscope.

Because of the high affinity of titanium for oxygen and nitrogen and the need to eliminate contamination variability, all welding and Varestraint testing was performed in an inert atmosphere glove box. The glove box was maintained at oxygen and nitrogen levels below 30 ppm.

All welds were sectioned in both the longitudinal and transverse directions, then ground and polished for metallographic examination using conventional techniques. Standard metallographic techniques were used to determine weld metal microstructure.

One weld specimen was Varestraint tested for each yttrium content studied, using four percent augmented strain. Cracks were identified and measured using a low magnification (15–60X) stereomicroscope.

To better understand the relationship between weld pool shape and bead morphology, decanting experiments were performed. A high-pressured argon blast was employed to expel the liquid weld pool. The decanting also allowed for determination of the weld pool volume. Decanted weld pools were filled with clay, which was then extracted, weighed and converted to a liquid metal volume. One weld pool was decanted for each yttrium content studied.

Because the grains in the fusion zone were highly elongated, the grain width was chosen as an index of grain refinement. Grain width measurements were performed on a weld representing each yttrium content, following ASTM E112-82 procedures.

For reasons which will be discussed later, a few static spot welds were made using the GTA welding torch. While the torch was held stationary, the arc was initiated, then extinguished after ten seconds. The molten pool solidified while under the torch gas. The welding parameters were the same as in all other welds except that the travel velocity was zero. The energy input for each static weld was 25.5 kJ.

Table 1—Chemical Composition of the Titanium Base Plate (wt-%)

Al	Nb	Ta	Mo	Fe	N	O	C	H
6.0	1.9	0.90	0.81	0.09	0.009	0.071	0.02	0.005

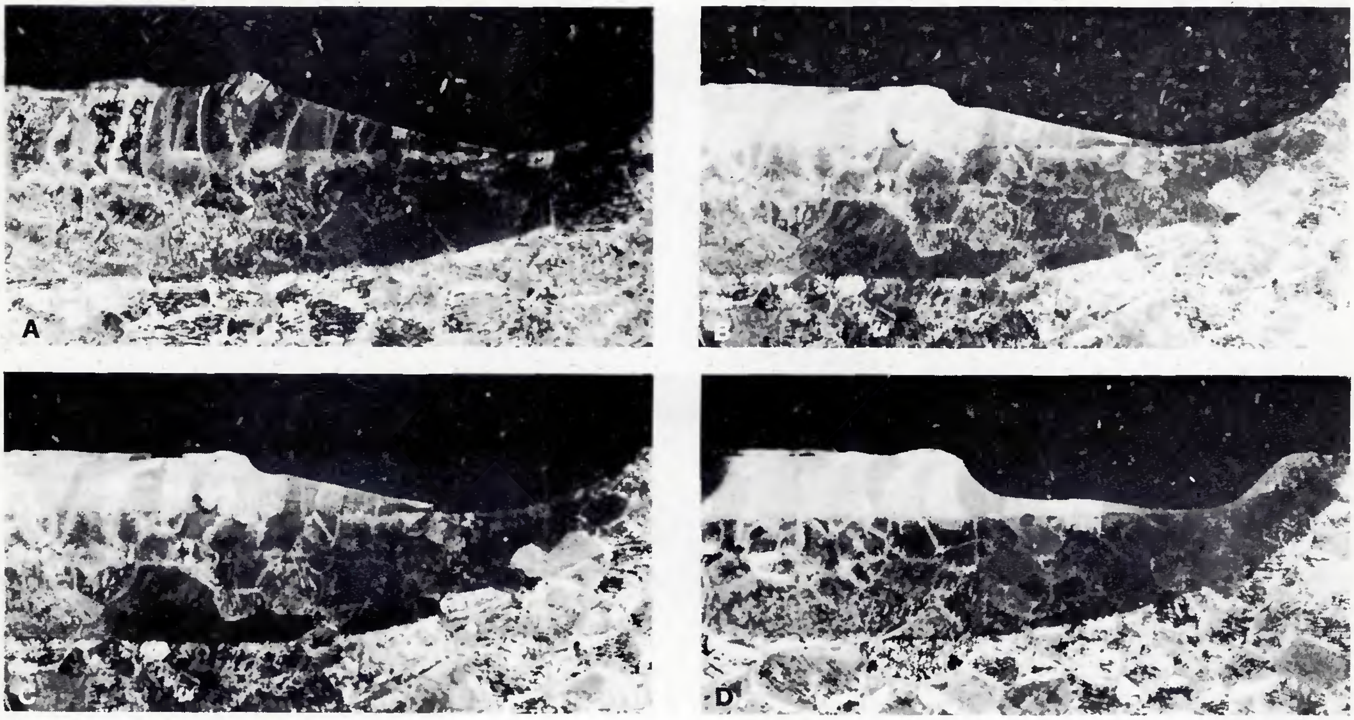

Fig. 5 — *Longitudinal sections of decanted weld pools. A — 0.0 wt-% Y; B — 0.02 wt-% Y; C — 0.08 wt-% Y; D — 0.20 wt-% Y*

Results and Discussion

Yttrium metal was introduced to Ti-6211 weld metal to achieve two main objectives. The first objective was to provide a mechanism by which the nucleation and growth processes during solidification of weld metal could be perturbed and, therefore, studied. The second objective was to investigate the potential for using yttrium additions to suppress or eliminate the cracking susceptibility of Ti-6211 weldments or to otherwise improve the weldment properties. Consequently, the results and discussion which follow will focus on, first, solidification kinetics and second, on a study of Varestraint cracking susceptibility.

Solidification Kinetics

In this investigation, large variations in weld penetration and bead morphology were observed as a result of small additions of yttrium (0.02 to 0.24 wt-%). Figure 5 shows longitudinal sections of decanted weld pools at various levels of yttrium content. It can be seen that a drastic change in bead morphology occurred when yttrium was added to the weld pool. Figure 5A shows a decanted weld pool containing no yttrium additions. In this case, the trailing edge of the weld pool tapered back at a constant slope from the fusion boundary, and the prior beta grains grew perpendicular to the advancing solid-liquid interface. In contrast, Fig. 5B shows a decanted weld pool containing 0.02 wt-% yttrium. In this weld pool, near the free surface, a small plateau developed, indicating that the direction of advance of the solid-liquid interface had changed. The growth direction of the prior beta grains remained perpendicular to the solid-liquid interface, resulting in a drastic change in grain orientation near the weld pool surface. Increasing the weld metal yttrium content increased the depth at which the plateau developed, as shown in Figs. 5C and 5D.

It is proposed that a change in fluid flow or an onset of turbulent flow was responsible for the observed change in weld pool shape. During welding when no yttrium was added, the weld pool appeared calm. As the yttrium content of the weld pool was increased, the surface of the weld pool became turbulent. Evidence of this change in the weld pool fluid flow was seen in the development of ripples on the fusion zone surface — Fig. 6. These ripples were seen on the as-welded surface, *i.e.,* no polishing or etching was done to reveal them. An important observation is that surface rippling (turbulent flow) was not detectable at 0.01 wt-% yttrium, but that additions of 0.02 wt-% yttrium or more resulted in surface rippling and a change in bead morphology.

Surface rippling has been attributed to surface tension gradients on the weld pool surface in GTA welding (Ref. 10) and during laser surface melting (Ref. 11). Therefore, it appears that yttrium altered the surface tension of the titanium alloy weld pool and affected the fluid flow in a way similar to that proposed by Heiple and Roper (Ref. 12).

Surface rippling has also been attributed to growth rate fluctuations during weld metal solidification (Ref. 13). To determine if the rippled surface was due to turbulent flow or growth rate fluctuations, static spot welds containing different levels of yttrium were made. The top surfaces of these welds are shown in Fig. 7. The weld shown in Fig. 7A contained no yttrium and appeared smooth, as expected. When yttrium was added, the fluid flow became so turbulent that liquid metal was forced over the edge of the weld pool, as indicated in Fig. 7B. This indicates that the surface ripples, as shown in Fig. 6, were due to the turbulent flow during the welding process.

There are two possible mechanisms by which yttrium could alter the surface tension and, thus, the fluid flow in the weld pool. One is that yttrium could remove oxygen by forming yttrium oxide. Oxygen is known to be surface-active and affect the weld pool fluid flow. This mechanism was proposed by Heiple, *et al.,* to explain the influence of aluminum in Type 21-6-9 stainless steel weldments (Ref. 14). In this titanium investigation, increasing amounts of yttrium added to the weld pool caused increasing turbulence. Obviously, the added yttrium should reduce the available surface-active free oxygen, and its effect should be to reduce turbulence, but this was not observed. This suggests that factors other than removal of surface oxygen contribute to weld pool turbulence. The other possible mechanism is that yttrium directly affected the temperature dependence

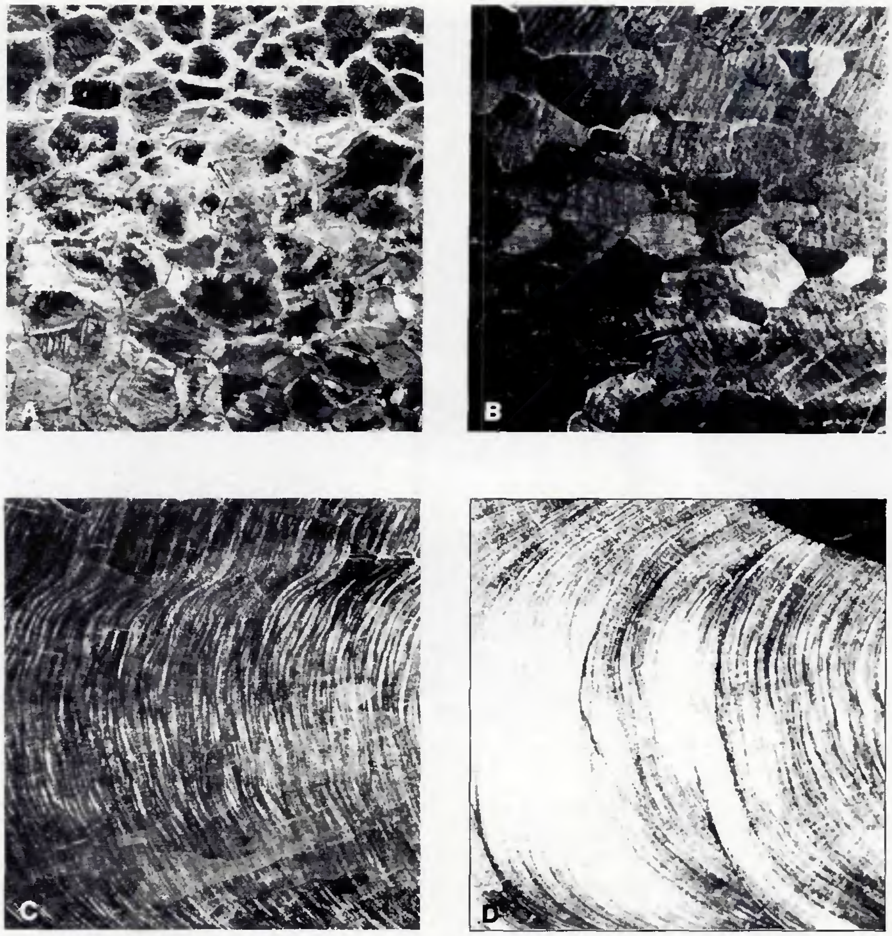

Fig. 6 — Macrographs showing fusion zone surface ripples. A — 0.0 wt-% Y; B — 0.02 wt-% Y; C — 0.08 wt-% Y; D — 0.20 wt-% Y

of the surface tension. Yttrium has been reported by Aledsondrov (Ref. 15) to alter the surface tension of molten steel. A comparison of the effects of yttrium, cerium and aluminum on the surface properties of molten steel confirmed that yttrium had the greatest influence. Yttrium could be a surface-active element in titanium, and thereby alter the surface tension directly. Although the mechanism for doing so is not entirely clear, the observations of this study indicate that yttrium additions altered weld pool fluid flow, producing outwardly directed flow at the weld pool surface from the center to the edges.

In addition to the change in bead morphology with increasing yttrium additions, shown in Fig. 5, a decrease in the depth of penetration was also observed, as shown in Fig. 8. Note that the fusion zone width remained constant. Further investigation revealed that the cross-sectional area of the fusion zone decreased as the yttrium content increased, as shown by Fig. 9. These observations indicate that the decreased penetration was caused by a reduction in the heat input available for melting.

Heat input is the quantity of energy introduced per unit length of weld from a traveling heat source such as an arc. In welding, the heat input is calculated by:

$$H_{net} = \eta \, VI/v \qquad (5)$$

where the efficiency, η, ranges from 0.21 to 0.48 for GTA welding (Ref. 7). Jackson and Shrubsall found that the cross-sectional area of a weld at constant travel speed was directly proportional to the heat input (Ref. 16). Hayduk (Ref. 17) showed that the fusion zone cross-sectional area of GTA autogenous welds made on Ti-6211 plate increased as the heat input was increased, as seen in Fig. 10.

For the observed decrease in cross-sectional area to occur as shown in Fig. 9, the effective or net heat input must decrease. In the heat input equation, both travel speed, v, and current, I, were held constant. Therefore, the efficiency of the heat source, η, or the voltage across the arc, V, decreased. To determine the effects of yttrium on the arc voltage, arc stability measurements were made. An oscilloscope was used to determine the potential across the arc and the stability of the arc potential. Neither the voltage, V, nor the voltage stability, ΔV, changed with the addition of yttrium. To further substantiate the fact that the heat input decreased, weld pool volume measurements were made. Weld pools containing various amounts of yttrium were decanted during the welding process, and the volume of the remaining weld pool cavity was then measured. The volume of metal melted is seen to decrease substantially with the addition of yttrium, as shown in Fig. 11. This observation indicates that either the potential across the arc was altered or the heat flow from the weld pool was enhanced. No change in the total potential across the arc was observed when yttrium was added to the weld pool. Evidently, increased turbulence of the weld pool, caused by the addition of yttrium, increased the heat transfer from the fusion zone and decreased the available heat.

The addition of yttrium was found to promote grain nucleation in the fusion zone of autogenous GTA Ti-6211 weldments, resulting in refinement of the grain structure, as shown in Figs. 12 and 13. In multipass welds, this grain refinement

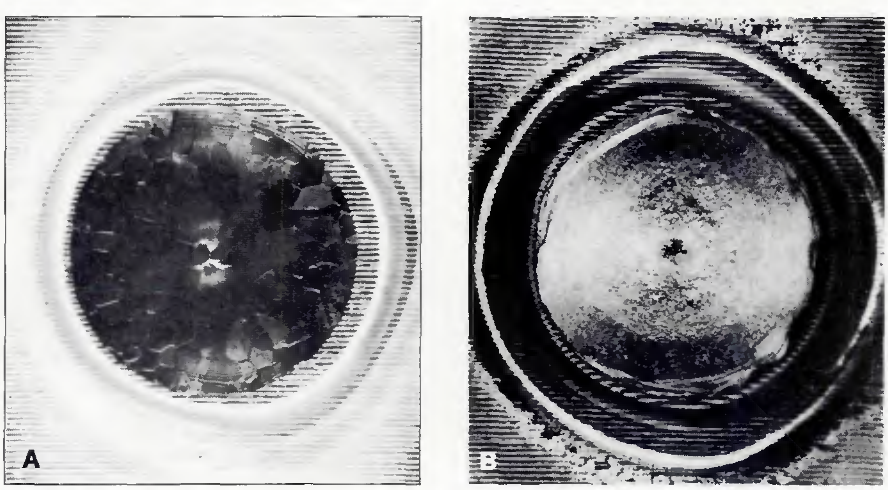

Fig. 7 — Static spot welds. A — 0.0 wt-% Y; B — 0.18 wt-% Y. Note turbulent flow around edge of pool in B

would definitely prevent epitaxial grain growth from bead to bead, thereby improving the weld metal properties.

Grain refinement in Ti-6211 welds, caused by yttrium additions, may relate to several fundamental mechanisms. These mechanisms are: 1) yttrium additions inhibit grain growth during solidification, thereby allowing more time for nucleation, 2) yttrium additions create fine oxide particles which act as nucleation sites for heterogeneous nucleation, 3) yttrium additions change the growth direction of grains in the fusion zone which in turn causes nucleation, 4) yttrium additions alter surface tension and fluid flow of the weld pool (Marangoni effect) and refine the structure by grain multiplication, and 5) yttrium additions alter the weld pool fluid flow and heat extraction from the pool, thereby increasing the degree of supercooling at the solid-liquid interface, which increases heterogeneous nucleation. Each of these grain refinement mechanisms will be discussed.

Effect of Yttrium on Growth Rate

If yttrium additions inhibit grain growth in the fusion zone during solidification, there would be more time for nucleation, and grain refinement would result. Solidification rates in weldments have been determined by many researchers (Refs. 18, 19). Figure 1 is a schematic diagram of the vector approach used to determine the solidification rates at different positions of the weld pool. This vector approach assumes that crystal growth is isotropic and takes place along the direction of the maximum thermal gradient. Thus, the lowest growth rates are found at the weld pool edge, where $\Theta = 90$ ($R = 0$), and the highest growth rate is at the weld pool centerline where $\Theta = 0$ ($R = \nu$). In this study, the important growth rate is not that of the trailing edge of the weld pool, but that of the grains growing from the fusion boundary to the top of the weld pool. Figure 1 is a schematic representation of a longitudinal section through a fusion zone, where R is the growth rate of the long columnar grains seen in titanium weldments. If the growth rate of these grains is slow enough, nucleation can occur before they grow across the fusion zone.

Longitudinal sections of yttrium-doped weldments were examined and growth rates were determined. Yttrium additions were found to have a large effect on the prior beta grain growth rate. Figures 14 (a schematic diagram) and 15 (actual macrographs) examine longitudinal sections through the fusion zone and show the effects of yttrium additions on the prior beta grain growth rate and morphology. Yttrium additions are seen to make the direction of grain growth more nearly

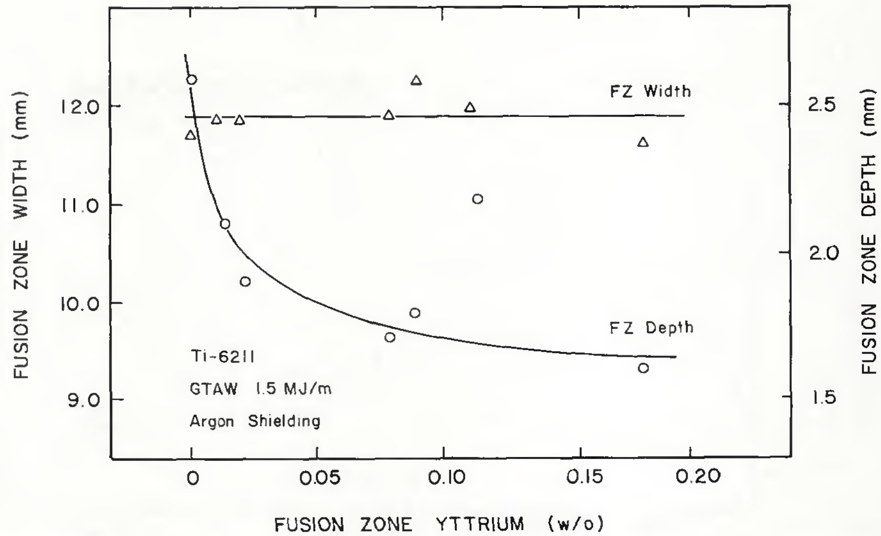

Fig. 8—Effect of yttrium on Ti-6211 fusion zone width and depth (penetration)

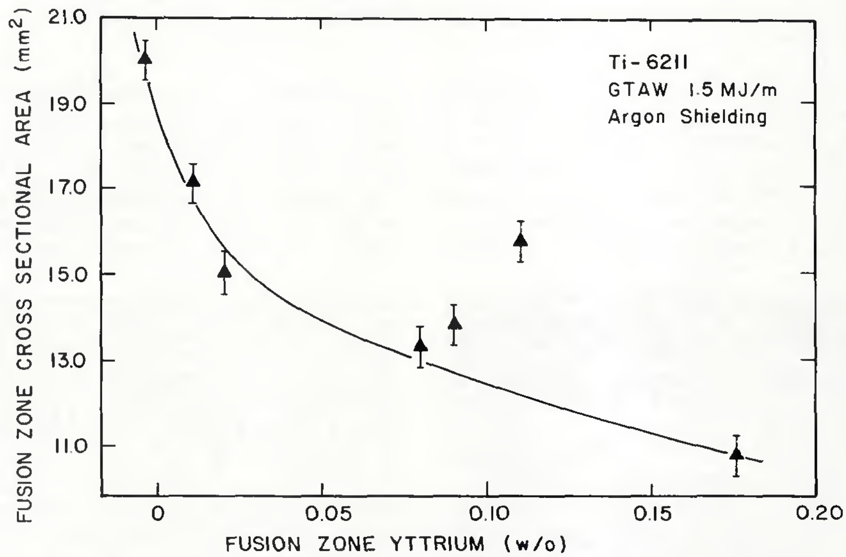

Fig. 9—Fusion zone cross-sectional area as a function of weight percent yttrium in the fusion zone

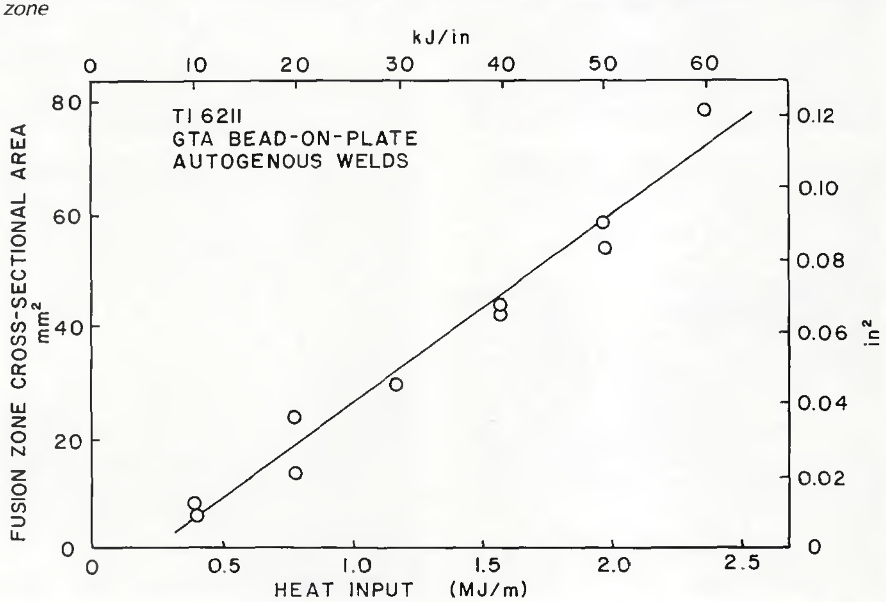

Fig. 10—Fusion zone cross-sectional area as a function of heat input for GTA Ti-6211 weldments (Ref. 17)

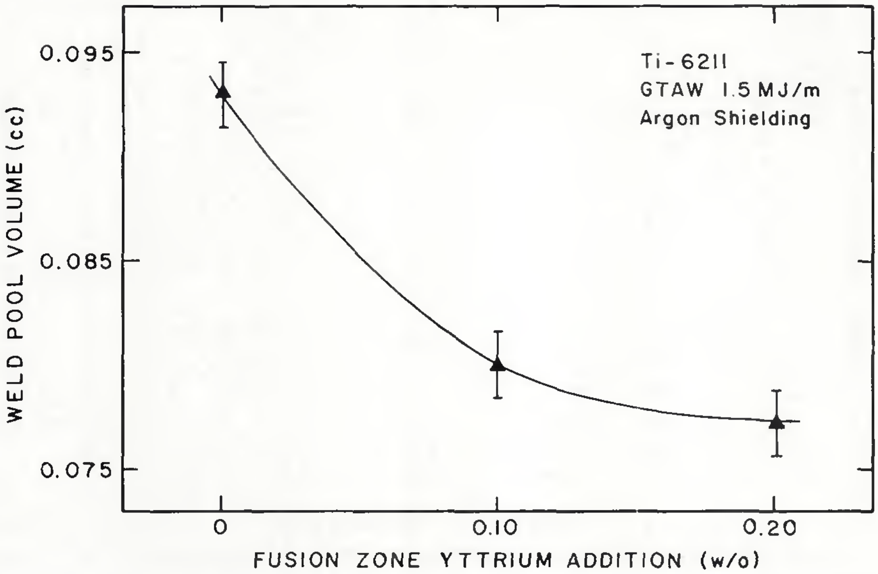

Fig. 11 — Weld pool volume as a function of fusion zone yttrium for autogenous Ti-6211 weldments

parallel to the weld axis (decreasing the angle Θ), a change which corresponds to an increase in growth rate. Figure 16 shows the growth rate (solidification velocity) of the prior beta grains versus the depth from the weld pool surface for various amounts of yttrium in the fusion zone. Yttrium increases the solidification rate in the fusion zone, such that there are two distinct regions of different growth rates, as seen in Fig. 15. The first region extends from the fusion boundary into the fusion zone. For weldments containing no yttrium, this region extends across the entire fusion zone. As yttrium is added, the extent of this region is reduced but the growth rate remains constant. The second region is near the surface of the fusion zone where Θ approaches zero and R approaches ν. This corresponds to a region with an accelerated growth rate. Increasing the weight percent yttrium increases the depth of this region below the weld pool surface.

To verify that yttrium was distributed uniformly throughout the fusion zone, microprobe scans were made across the fusion zones to determine any variation in yttrium concentration. Yttrium content as a function of depth from the surface of the fusion zone is shown in Fig. 17 (these data are from the same welds analyzed in Fig. 16). A uniform distribution of yttrium throughout the fusion zone was observed. This observation indicates that the change in grain morphology and growth rate within the fusion zone does not result from insufficient mixing or segregation. The addition of yttrium had an indirect effect on growth rate. If yttrium increased the growth rate directly, a constant and unique growth rate would have been observed for each yttrium concentration studied.

The postulate that yttrium additions inhibit grain growth in the fusion zone during solidification and cause grain refinement clearly is incorrect. Yttrium additions increase the growth rate and, in the absence of other effects, would cause grain coarsening.

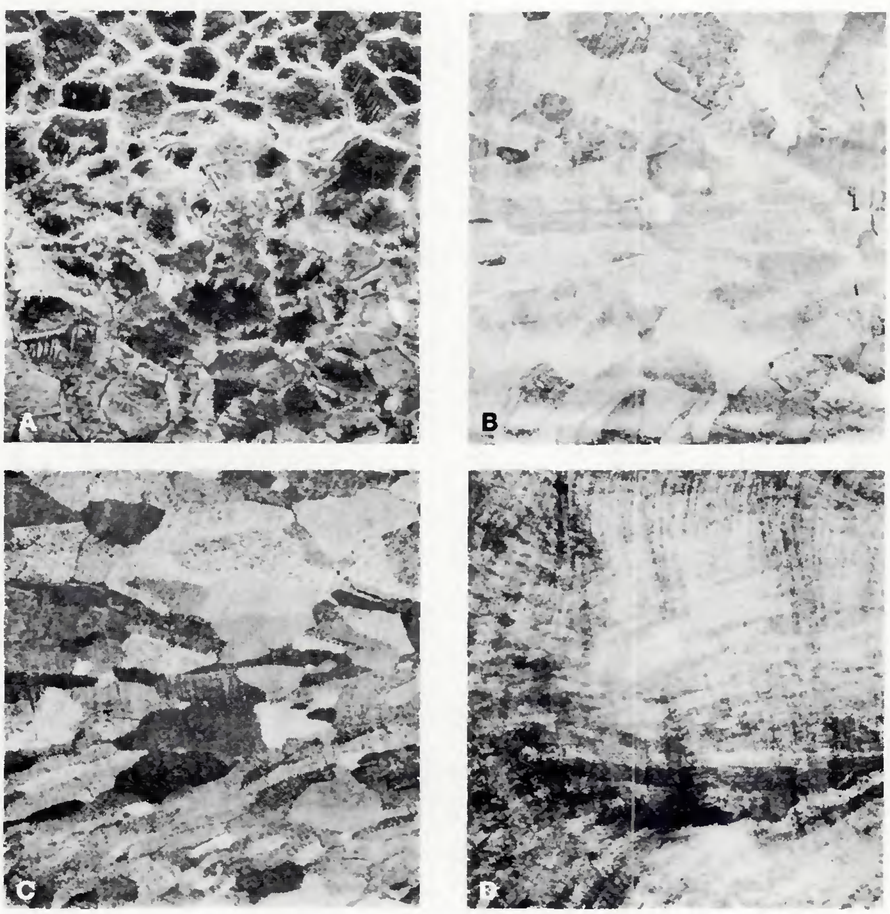

Fig. 12 — Effect of yttrium on Ti-6211 fusion zone grain size. A — 0.0 wt-% Y; B — 0.02 wt-% Y; C — 0.09 wt-% Y; D — 0.24 wt-% Y

Effect of Yttrium as an Inoculant

The introduction briefly described previous attempts to refine the grain structure of titanium weldments by inoculation (Refs. 5–7). Yttrium was selected because of its low solid solubility in titanium and its high affinity for oxygen relative to titanium (Ref. 9). The yttrium forms yttrium oxide particles in the fusion zone. The point to be considered is whether or not the yttria particles act as high-melting-temperature inoculants.

In this study of yttrium effects on titanium weldments, and in the study of Misra, *et al.* (Ref. 6), a uniform distribution of Y_2O_3 second-phase particles was observed throughout the fusion zone. The grain refinement resulting from yttrium additions can be seen in Fig. 15C. In this longitudinal section of the fusion zone, grain refinement does not occur uniformly across the fusion zone. If the uniformly distributed yttria particles acted as effective inoculants for heterogeneous nucleation, the grain refinement would have been uniform. Consequently, the primary cause of grain refinement in yttrium-doped titanium weld metal is not inoculation of the melt by yttrium-oxide particles. Nucleation probably occurs heterogeneously on yttria particles, but growth is dependent upon favorable thermal conditions.

Nucleation at Position of Thermal Gradient Reorientation

Initial solidification at the fusion boundary occurs epitaxially. Grains that grow epitaxially tend to grow along certain preferred crystallographic, or easy growth, directions. In body-centered cubic metals, this is the <100> direction. The driving force for solidification is the thermal gradient, which defines the direction of preferred grain growth. Grain growth is favored for those initially nucleated grains having their easy growth directions parallel to the maximum thermal gradient. These grains grow at the expense of less favorably oriented grains. When the direction of the maximum thermal gradient in the weld pool changes, it is necessary for crystals to change orientation during growth if they are to continue to follow this maximum gradient. Alternatively, nucleation of new grains may occur at a position in which there is a drastic reorientation of the thermal gradient. The new grains then grow along a <100> direction parallel with the new thermal gradient.

Yttrium additions were found to change the growth direction of grains in the fusion zone, as previously discussed and shown in Fig. 15. This change in orientation during growth could be responsible for the nucleation of new grains. However, two observations indicate that nucleation of columnar grains did not occur because of reorientation. First, it can be seen in Fig. 15B that the existing grains changed growth direction near the surface without the nucleation of new grains. Second, it was also observed that nucleation of new grains

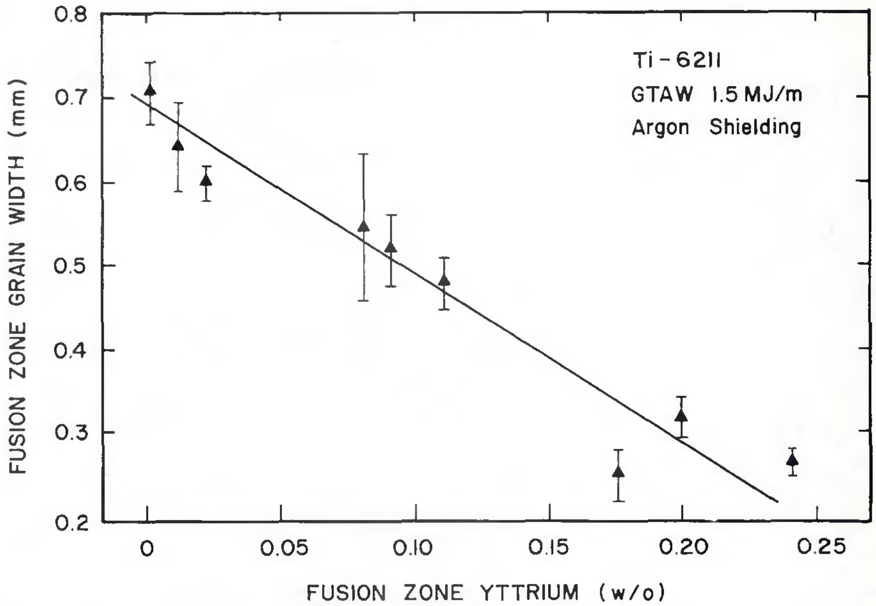

Fig. 13 — Effect of weld metal yttrium content on fusion zone grain width

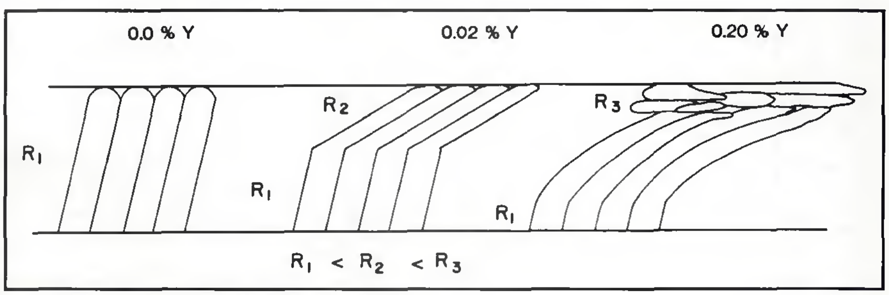

Fig. 14 — Effect of yttrium on growth rates in GTA Ti-6211 weld metal

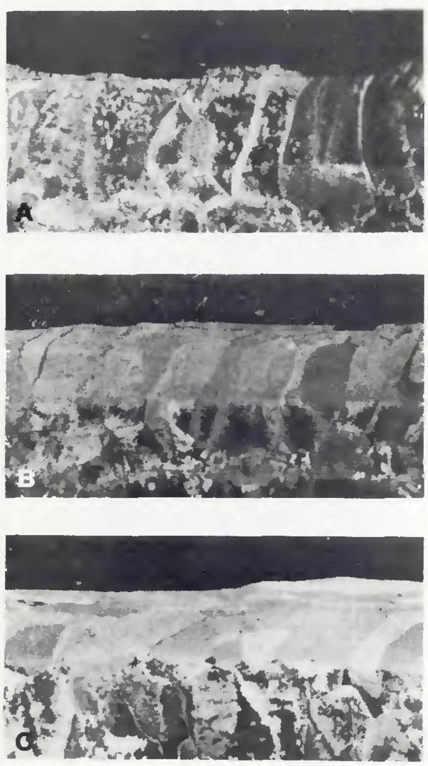

Fig. 15 — Macrographs of longitudinal sections through the fusion zone showing change in grain morphology and growth rate. A — 0.0 wt-% Y; B — 0.02 wt-% Y; C — 0.24 wt-% Y

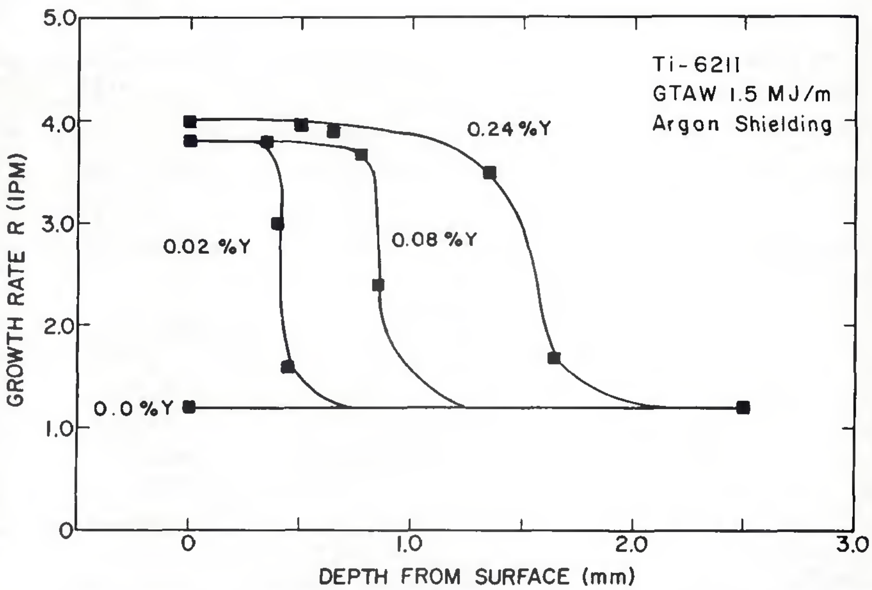

Fig. 16 — Growth rate, R, as a function of the depth from the fusion zone surface for various weld metal yttrium contents

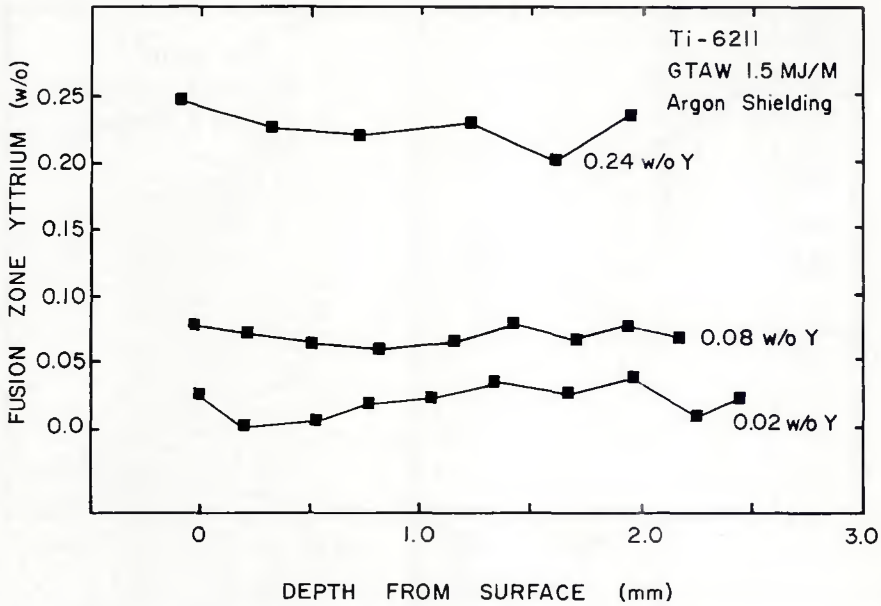

Fig. 17 — Yttrium content as a function of depth from the surface of the fusion zone for weldments in Fig. 16

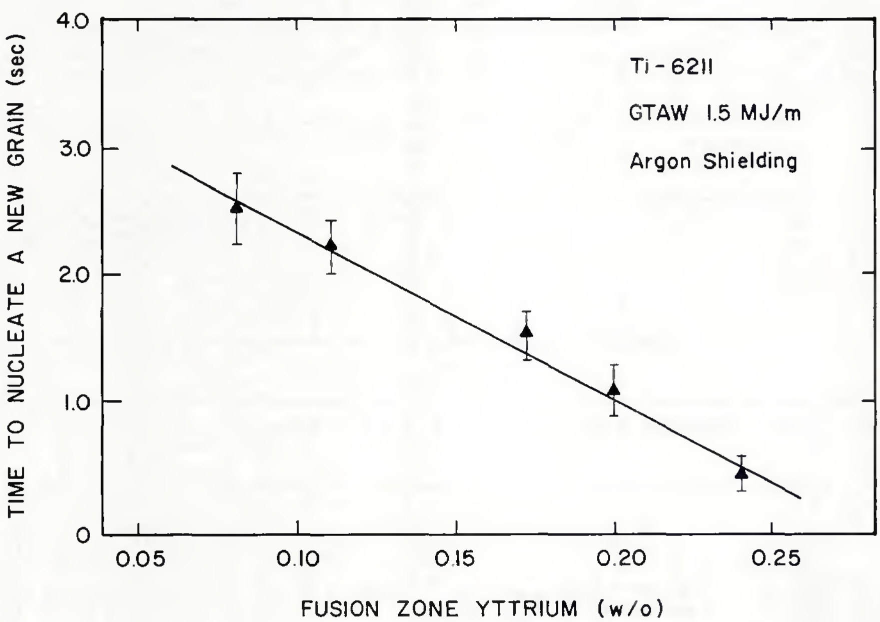

Fig. 18 — Time to nucleate a new grain as a function of the fusion zone yttrium content

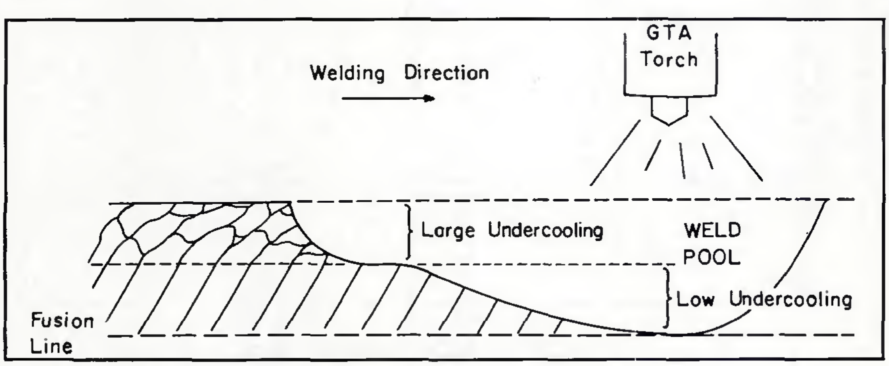

Fig. 19 — Longitudinal section of weld pool and grain structure, showing regions of different undercooling and resulting nucleation

did occur where grains were growing parallel to the welding direction with no reorientation. To further eliminate the change in growth direction as a possible rate-controlling nucleation mechanism, static spot welds (*i.e.*, torch velocity of zero) were made to eliminate the dynamics of a moving heat source. Three welds containing no yttrium, low yttrium (0.1 wt-%), and high yttrium (0.2 wt-%) were allowed to solidify under the cover gas of the torch. It was found that yttria-enhanced nucleation and the consequent grain refinement occurred even when there was no change in the direction of grain growth.

Grain Refinement and Weld Pool Fluid Flow

The effect of yttrium on turbulence and fluid flow has already been discussed. Various researchers have proposed that convection (Ref. 9) or electromagnetic stirring (Refs. 19–21) of the weld pool can cause grain refinement when dendrite fragments, melted or broken off from the solidifying interface, nucleate new grains. However, in the weld pool, the temperatures are very high because of the localized heat source. This makes survival of the dendritic nuclei difficult, and grain multiplication by this mechanism is unlikely (Ref. 19). However, alterations in weld pool fluid flow may perturb the heat transfer, thereby altering the solidification kinetics. These potential effects will be subsequently discussed.

Effect of Yttrium on Undercooling

Changes in the degree of undercooling at the solid-liquid interface should influence grain refinement. Since nucleation is enhanced by increased undercooling, any increase in undercooling would make yttria particles more effective nuclei. In the nucleation and growth model developed by Misra, *et al.* (Ref. 6), it was assumed that the weld pool undercooled as a bulk liquid. Based on this assumption, yttrium was postulated to reduce the time for nucleation. This nucleation time dependence explained why grain refinement occurred near the center and the surface of the weld, and not near the fusion boundary. In the present study, it was assumed that the critical undercooling responsible for nucleation was near the solid-liquid interface and not in the bulk of the liquid. The amount of undercooling within the supercooled region is dependent on the actual temperature gradient at the solid-liquid interface and the profile of the liquidus temperature. If undercooling at the interface increased, a yttria particle within this undercooled zone would become a more effective nucleation site, and heterogeneous nucleation could occur. The amount of undercooling could be increased either by increasing the thermal gradient (increasing the cooling rate) or by altering the liquidus temperature. Yttrium additions could possibly increase the thermal gradient within the liquid. Yttrium has been shown to induce turbulent flow in the weld pool, which might enhance heat transfer from the pool and increase the cooling rate. This increased cooling rate would then increase the thermal gradient at the solid-liquid interface, and hence, the undercooling.

The actual nucleation times, t_n (the time required to nucleate a new grain during solidification), were found in this study to decrease with increasing yttrium content, as shown in Fig. 18. The decrease in t_n is believed to result from a change in the location of the highly undercooled liquid. Decanted weld pools showed that the development of a plateau region, previously shown in Fig. 5, became more pronounced with increasing yttrium content. We have suggested that a change in fluid flow caused this plateau to develop, and it seems plausible that the change in fluid flow resulted in increased undercooling. Enhanced nucleation and grain refinement were observed to occur in the weld metal at or above the region corresponding to this plateau, also suggesting that increased undercooling occurred at or above the plateau region. This concept is schematically illustrated in Fig. 19. When yttrium additions were increased, the depth of the plateau increased and the depth for highly undercooled liquid also increased. The result was an apparent decrease in the time to nucleate a grain, t_n, although t_n was actually a measure of the time for a grain to reach the region of increased undercooling.

Cracking Susceptibility of Ti-6211

The high cracking susceptibility of Ti-6211 weld metal has been related to the loss of ductility at elevated temperatures (Ref. 17). This high-temperature ductility loss is a concern in multipass weldments where material is exposed to thermal cycles in the presence of residual stress.

Rath, *et al.* (Ref. 22), investigated the influence of erbium and yttrium additions on the microstructure and mechanical properties of titanium alloys. They found that alloying additions of yttrium and erbium significantly retarded recrystallization and grain growth in alloys subjected to elevated temperatures. Seagle, *et al.* (Ref. 23), studied the effect of yttrium on the microstructure of Ti-6Al-4V. They found that small additions of yttrium, to 0.20 wt-%, reduced grain growth in Ti-6Al-4V heated above the beta transus. Sastry, *et al.* (Ref. 24), studied the effect of yttrium-oxide and metallic-yttrium additions on the room-temperature fracture toughness and temperature deformation characteristics of Ti-6211. From elevated temperature compression tests of as-cast and forged specimens, they found that yttrium and yttria additions in concentrations of less than 0.012 wt-% reduced cracking during forging of Ti-6211. They also discovered that yttrium and yttria additions slightly decreased the fracture toughness.

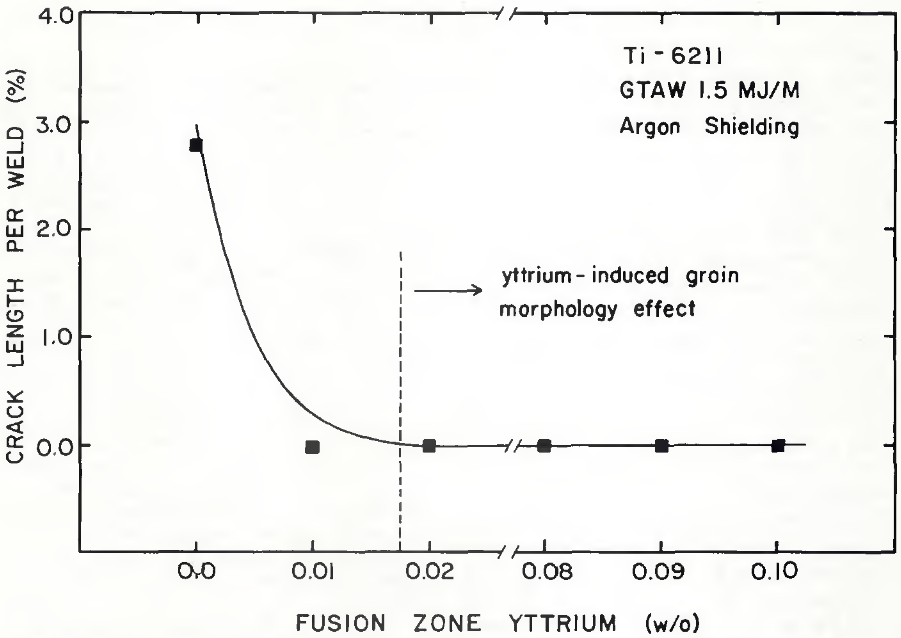

Fig. 20 — Varestraint fusion zone cracking versus weight percent yttrium for Ti-6211 GTA weldments

In the work by Edwards, Olson and Matlock (Ref. 5), the influence of small additions of yttrium, erbium and zirconium on the microstructure of Ti-6211 was studied. They showed that these microalloying additions reduced the prime lath length in the weld metal, with yttrium additions being the more effective. Lewis (Ref. 25), in his study of the hot ductility of Ti-6211, investigated the effect of yttrium additions in Ti-6211 plate. He found that Ti-6211 plate containing 150 ppm yttrium, added as Y_2O_3, exhibited significantly higher hot ductility (23 to 45% reduction in area) than the yttrium-free alloy (0 to 15% reduction in area). Samples of alloys were thermally cycled and tested in a manner expected to produce low hot ductility. Lewis proposed that the hot ductility loss in Ti-6211 caused by impurity (probably sulfur) segregation to the grain boundaries could be reversed by yttrium additions that scavenge the embrittling impurities, thereby increasing prior beta grain boundary strength.

As part of this current study, Varestraint testing was performed on Ti-6211 fusion welds containing various amounts of yttrium. The cracking observed was intergranular and was similar in all respects to the high-temperature cracking observed by Lewis (Ref. 25) and by Hayduk, *et al.* (Ref. 26). Metal microalloyed with 0.01 to 0.10 wt-% yttrium did not exhibit any loss in ductility when "varestrained." This is illustrated in Fig. 20, a plot of Varestraint fusion-zone cracking versus weight-percent yttrium in the fusion zone. The amount of cracking in the fusion zone was measured on the surface of the weldment and not on longitudinal or transverse sections.

Bowden and Stake (Ref. 27) found that the ductility at high temperature in Ti-6211 was improved when the elongated beta grains were oriented so that their length was parallel to the applied tensile stress (Ref. 27). Beta grains of yttrium-doped Ti-6211 weld metal were found to be elongated parallel to the length of the weld bead (which is the axis of applied tensile stress in the Varestraint test), whereas the beta grains in yttrium-free weld metal were nearly perpendicular to the applied stress. The effect of grain morphology on the cracking susceptibility of Ti-6211 weld metal is thus similar to that observed by Bowden and Starke for Ti-6211 base plate.

The fusion zone grain morphology was altered by yttrium additions greater than 0.01 wt-%, as shown in the previous section, so that weldments containing greater than 0.01 wt-% yttrium were not susceptible to cracking. However, weld metal containing only 0.01 wt-% yttrium was also found to be resistant to Varestraint cracks (see Fig. 20), yet the beta grains were oriented with their length virtually perpendicular to the applied tensile stress axis. Cracking resistance in weldments containing only 0.01% yttrium must therefore result from a mechanism other than grain morphology effects.

Yttrium additions were found to refine the solid-state transformation products.

An observation of reduced cracking susceptibility corresponding to microstructural refinement is consistent with the observations of Hayduk, *et al.* (Ref. 26). That study showed that Ti-6Al-4V did not crack from Varestraint testing, and that Ti-6Al-4V had a much finer fusion zone microstructure than the Ti-6211 fusion zone weld metal. Hayduk, *et al.*, also reduced cracking in Varestraint-tested Ti-6211 weldments by increasing the weld metal cooling rate to create a refined fusion zone microstructure. The evidence suggests that yttrium particles may act as heterogeneous nucleation sites for the alpha phase, resulting in a refinement of the solid-state transformation products. It may also be possible that an enhanced cooling rate, caused by an yttrium-induced change in weld pool fluid flow, persists, affecting the solid-state transformations. Clearly, much remains to be learned regarding the effects of yttrium on the microstructure of titanium weld metal.

Conclusions

1) Yttrium additions made to Ti-6211 weld metal alter bead morphology by changes in the weld pool fluid flow and heat extraction.

2) Yttrium additions promote nucleation in the fusion zone, resulting in refinement of the as-solidified grain structure.

3) Increases in the heterogeneous nucleation rate are most likely caused by increases in the undercooling at the solid-liquid interface during solidification. These increases result from changes in fluid flow and heat extraction.

4) Yttrium additions of the weld pool reduce the fusion zone Varestraint cracking susceptibility of GTA Ti-6211 autogenous weldments.

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