EFFECT OF WELD ENERGY INPUT AND POST-WELD HEAT-TREATMENT ON STRUCTURE AND PROPERTIES OF Ti-6A1-4V

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INTRODUCTION

The alpha-beta Ti-6Al-4V alloy is extensively used for a variety of components not only for aerospace applications but also in a number of other (e.g. chemical, marine, etc.) industries. Like other titanium alloys, it displays high strength-to-density ratio and excellent corrosion resistance. Both for considerations of design rationalisation and economics, welding is required for fabricating many of the above components.

Though the alloy Ti-6Al-4V is generally considered more easily weldable than other titanium alloys, the weld thermal cycle does indeed cause drastic microstructural changes which result in poor mechanical properties, particularly ductility. In the as-welded condition. This is primarily due to two factors, viz., a large prior-beta grain size and the formation of the martensitic alpha prime phase. Obviously, the former is promoted by large heat inputs and the latter by rapid cooling. If the heat input is reduced for limiting beta grain growth, the cooling rate will increase resulting in a greater percentage of martensite. Thus the two problems leading to low ductility can not be solved simultaneously by modifying weld energy input alone. Preheating has sometimes been suggested, but has received little attention (1) in view of practical difficulties due to the chemical reactivity of titanium. It is, therefore more rational to decrease energy input during welding and resort to a post-weld heat treatment that would convert the as-welded martensite to more ductile products.

Several options are available for post weld heat treatment. A stress-relieving and stabilizing treatment at about 600°C has been commonly used for Ti-6Al-4V weldments. Heat treatments at temperatures higher than 600°C have also been tried by other investigators (1,2). However, a temperature of 760°C apparently represents the maximum post-weld heat treatment temperature for the majority of welded aerospace structural components (3). It has generally been found that thermal treatments at still higher temperatures close to but below the beta-transus are often required before significant improvements in ductility can be achieved. An elaborate three-step heat treatment at 732, 927 and 732°C has been suggested for the alloy Ti-6Al-6V-2Sn (4) and successfully tried for Ti-6Al-4V (1). Nevertheless it has been subsequently demonstrated by baseslack (5) that a single-cycle heat treatment, which employs only the high-temperature portion of the triplex treatment, is equally effective in modifying microstructure and restoring ductility.

The present paper deals with the macrostructural and microstructural changes in the fusion zone (FZ) in Ti-6Al-4V sheet material as the weld heat input is altered by using various welding processes and as the welded joints are subjected to heat treatment at different temperatures. These changes are systematically corelated with the mechanical properties, viz., hardness, strength and ductility.

Experimental details

Materials

The material used for the investigation was an alpha-beta processed Ti-6Al-4V alloy of 6 mm thickness which was received in the mill-annealed condition. The chemical composition (weight %) was as follows :

Al - 6.01; V - 4.01; C - 0.01, S - 0.0005; Cr - 0.143; Fe - 0.125; H - 0.0075; O - 0.1140; N - 0.0042; Ti - Balance.

Welding

Three welding processes were used :-

* Manual gas tungsten-arc welding (M-GTAW) in a hemi-spherical 'glove box; of 1.5 metres diameter in which a positive pressure of high-purity (99.95%) argon was maintained

* Automatic gas tungsten-arc welding (a-GTAW) using auxillary trailing and backing gas shields in addition to the torch gas.

* Election beam welding (EBW). The welding parameters are given in details in Table 1.

Autogenous full-penetration bead-on-plate welds were made in all cases. In GTAW this required two passes to be deposited one on each side. Just prior to welding the base metal coupons were pickled in a 25 HNO_3 + 10 HF + 65 H_2O solution and degreased with acetone.

Heat Treatment

The welded strips were subjected to post-weld heat treatment at two different temperatures; 700°C and 900°C. The time of heat treatment was three hours which was followed by furnace cooling.

Earlier reports have indicated that a high-temperature heat treatment is required for ensuring adequate ductility. However, the fact that Ti-6Al-4V shows a beta-transus of 980°C sets an upper limit for the heat treatment. Furthermore, the alloy exhibits superplastic behaviour at about 950°C (1). In the current investigation, therefore, post-weld heat treatment was carried out at 900°C to derive the expected optimum benefit in mechanical properties and at 700°C to follow the structural changes and to see if it would sufficiently imporve the as welded condition. All heat treatments were performed in vaccum

TABLE 1 : WELDING PARAMETERS FOR MANUAL GTA, GTA AND ELECTRON-BEAM WELDING OF TI-6AL-4V

Process voltage	*Arc/beam voltage (V)*	*Arc/beam current (A)*	*Travl speed (mm/min)*	*Heatinput (j/m)*
Electron beam welding	100 X 10^3	63 X 10^3	1200	315
Automatic GTAW	12	200	300	480
Manual	14	220	275	672

Other parameters

Electron beam welding	:	Vacuum of 10^{-3} torr	Automic GTA	Torch gas flow rate = 71/min Trailing gas flow rate = 14 l/min Backing gas flow rate = 51/min
Manual GTAW	:	Torch gas flow rate = 15 l/min		

TABLE 2 : FUSION ZONE TENSILE PROPERTIES OF TI-6Al-4V WELDMENTS

Process	Conditions	Yield strength (MPa)	UTS (Mpa)	Elongation (%)
Electron	As-welded	894	1005	9.5
Beam	PWHT at 700oC	884	997	8
Welding	PWHT at 900oC	805	936	12
Automatic GTAW	As-welded	880	975	8
	PWHT at 900°C	784	896	12.5
Manual GTAW	As-welded	873	986	8.5
	PWHT at 700°C	866	965	6.5
	PWHT at 900°C	800	925	12

Note : (1) All results are an average of 2 tests each
(2) For compairason, the base metal properties were : 857MPa(Ys), 973MPa(UTS) and 14.5%(elongation).

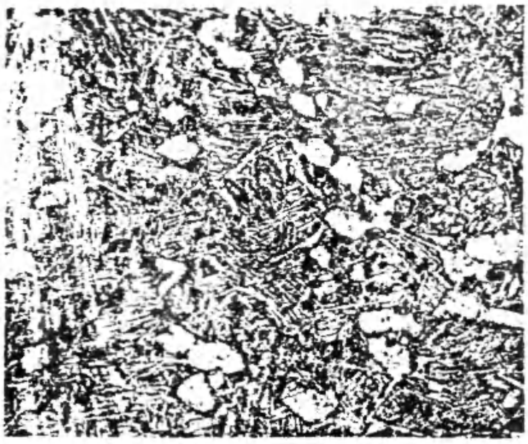


Fig 1 Microstructure of base metal (500 X)

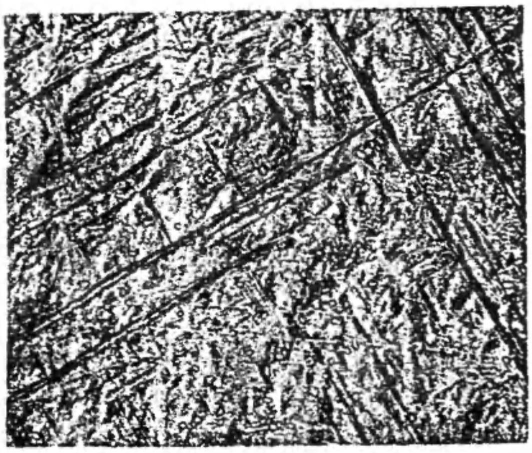


Fig : 4 Microstructure of base metal water-quenched from beta-phase field (500 X)

the furnace pressure being maintained at 10^{-5} torr.

Evaluation

Metallographic investigation was carried out using light microscopy. The study of macrostructures in alpha-beta titanium alloys is difficult because solute segregation effects acompanying solidification are partially masked by diffusion occurring during cooling in the beta-phase field and by the subsequent solid-state transformation (6). In the present study macroexamination was carried out using an etchant containing HF HNO_3 and special additives. For microstructural study the etchant was a 2 HF + 3 HNO_3 + 95 H_2O solution.

Tensile testing was conducted using longitudinally-oriented weld tensile specimens, i.e. the weld length alligned along the loading axis. Vickers hardness measurements were made across transverse sections of weldments in all welded and post-welded heat treated conditions using a 5 kg load. Tensile testing was performed on an instron machine at a cross-head speed of 0.5 mm/min. The results are given in Table 2

Results and Discussion

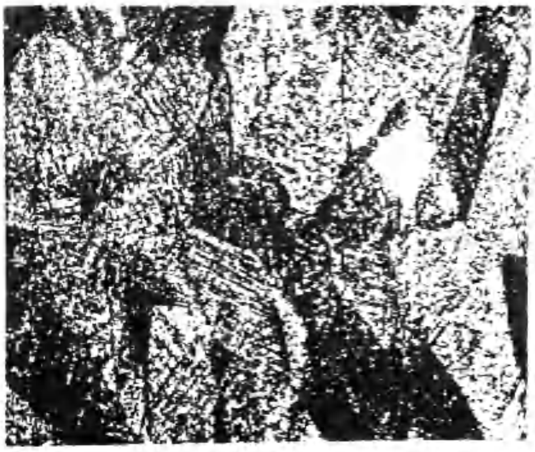
Structural Changes

The base metal microstructure given in Fig. 1 shows a large amount of primary alpha present with regions of transformed beta. During welding, locations in the heat-affected zone are heated to various sub-transus and super-transus temperatures before cooling to room temperature. In the fusion zone, solidification as beta crystals is followed by beta decomposition which may occur martensitically or by nucleation and growth depending on cooling rate.

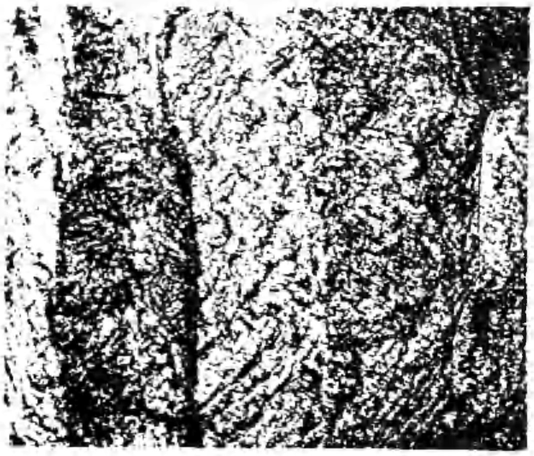
As-Welded Condition

The mactostructures of the weld metals produced by the three processes are given in Fig. 2. In all cases solidification is seen to occur in the form of large columnar beta grains, though this is partially broken up in the GTA welds because of the use of two passes and the consequent overlapping of beads. The macrostructures clearly show the increase in grain size as the heat input is increased progressively from EBW through A-GTAW to M-GTAW.

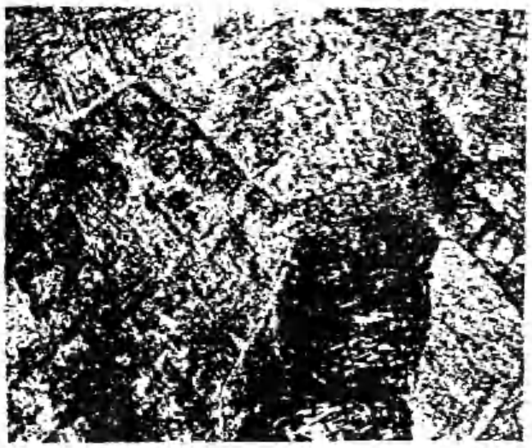
The FZ microstructures are seen in Fig. 3. The rapid cooling rate in EBW produces a fine acicular microstructure that is presumably entirely martensitic. For comparison, Figure 4 shows the microstructure of a specimen water-quenched from the beta-phase field, which is known to be fully martensitic. It is reasonable to presume that the high cooling rate in EBW would miss the nose of the CC1 diagram for the alloy and thus suppress any diffusion - based transformation.



a) Electron beam weld

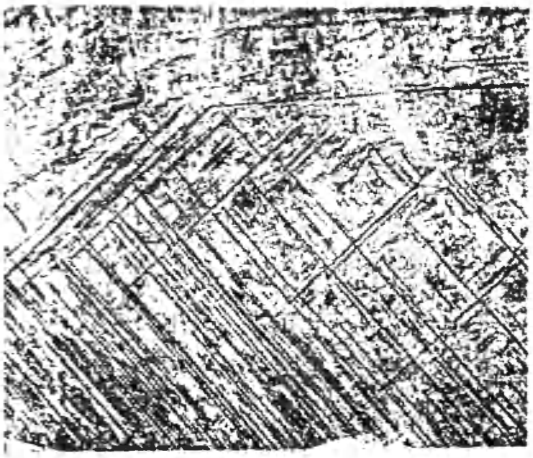


b) Automatic GTA weld

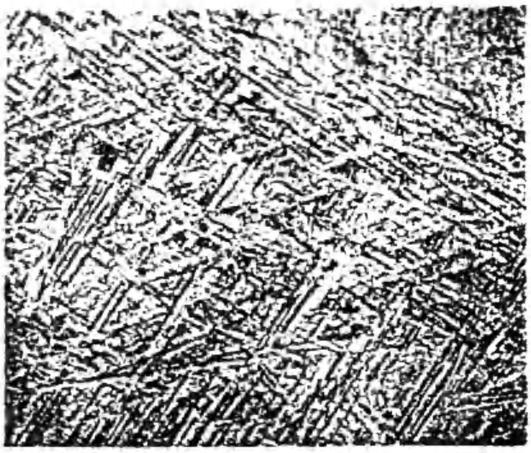


c) Manual GTA weld

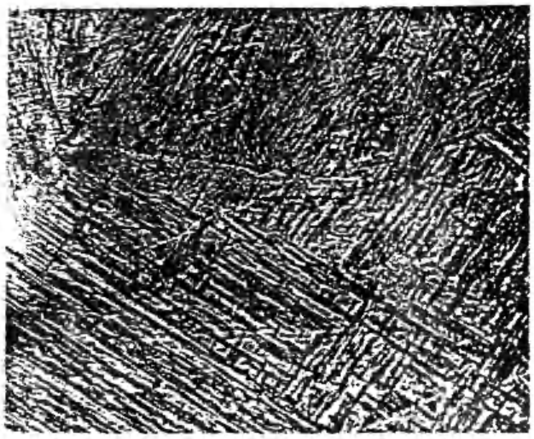
Fig 2 : Macrostructures of the fusion zone (50X)



a) Electron beam weld



b) Automatic GTA weld



c) Manual GTA weld

Fig 3 : Microstructures of the fusion zone as welded condition (50X)

As the heat input is increased, first in automatic GTAW and even more in manual GTAW, a general coarsening of the acicular intragranular microstructure may be observed, Fig. 3. The needles which are extremely fine in electron beam welds become somewhat thicker and less sharply defined in GTA welds. This is probably due to the greater percentage of the diffusional alpha-phase forming in GTAW. It is obvious that, as the cooling rate becomes smaller in GTAW, the degree of undercooling is low and the driving force for beta decomposition is small. Under such circumstances, the beta phase initially transforms by nucleation and growth in those regions where preferred nucleation sites exist, e.g., beta grain boundaries. This diffusional transformation precedes the martensitic reaction that occurs subsequently.

Post-Weld Heat treated Condition

The fusion zone microstructures after post-weld heat treatment at 700°C are shown in Fig. 5 and after treatment at 900°C in Fig. 6. These structures can be explained by considering how the metastable constituents of the weld metal microstructures transform on subsequent heat treatment.

During post-weld heat treatment two important reactions occur :-

* Decomposition of alpha-prime martensite to alpha and beta
* Precipitation of alpha in the retained metastable beta.

The transformation of alpha-prime has been shown to occur by heterogeneous nucleation and growth of the beta phase primarily of the martensite plate boundaries. This precipitation reduces the concentration of beta stabilizers in the supersaturated martensite until its composition approaches that of equilibrium alpha at the aging temperature (7). These aging phonomena start occuring even at a temperature of about 500°C and therefore could proceed to a considerable degree during heating to the post-weld heat treatment temperature. Subsequent overaging during the execusion and at heat treatment temperature promotes a coarsening of favourably-sized and oriented alpha plates at the expenses of the smaller, less fabourably-oriented plates (5).

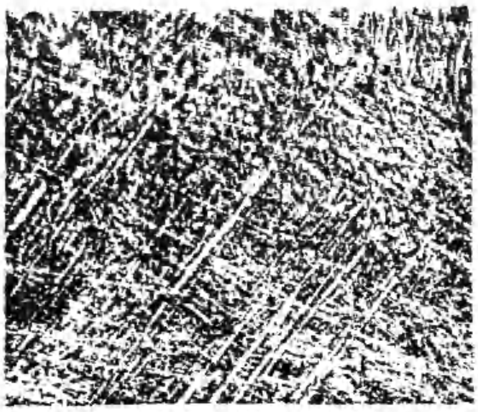
The FZ microstructures after post-weld heat treatment at 700°C show a very fine alpha-beta structure with little evidence of coarsening. However, as the post-weld heat treatment temperature is increased to 900°C, appreciable intragranular coarsening occurs while the lath appearance is still retained. Fig. 6, shows lenticular alpha plates densely packed in a beta matrix with no indication of any globularisation. An important feature of the 900°C heat treated microstructures is the appearance of the alpha phase at the prior-beta grain boundaries. This is at variance with an earlier investigation by Borgreen and Willson (1) in which GTA-welded Ti-6Al-4V sheet did not show any grain boundary alpha on post-weld heat treatment at 900°C. In our investigation, the 700°C heat treatment did not result in any discernible grain boundary alpha, but the treatment at 900°C has clearly led to alpha precipitation at the crystal boundaries. However, even at 900°C the alpha phase is not continuous along the grain boundaries. It may be incidentally noted that the post-weld heat treatment has not changed the prior-beta grain size.

It may also be seen that the heat treatment at 90C°C has resulted in nearly indentical microstructures in all the three cases eliminating the differences in the as-welded condition caused by changes in welding heat input.

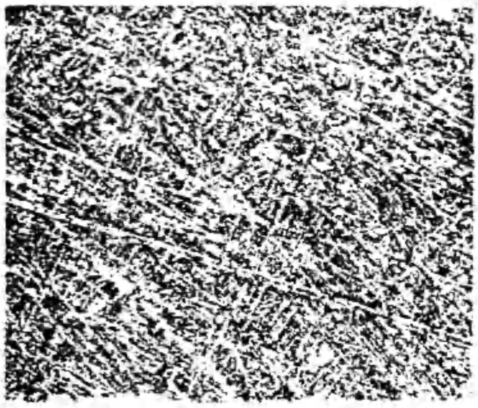
Mechanical Properties

The results of the tensile tests are given in Table 2. The base metal has a yield strength of 857 MPa, UTS of 973 MPa and tensile elongation of 14.5%. It is seen that welding leads to a reduction in fusion zone ductility in all cases, while the strength values remain substantially unaltered. The reduction in ductility is not doubt due to the large prior-beta grain size (Fig. 2) and the predominantly acicular microstructure (Fig. 3).

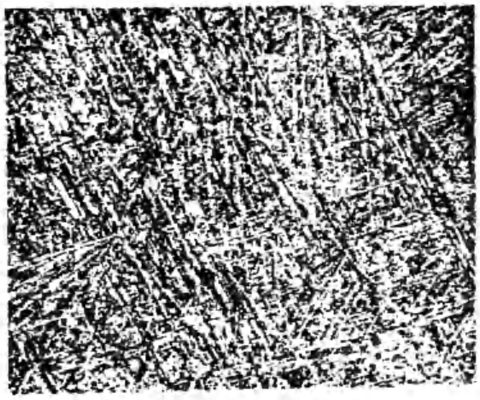
A comparison of Figure 2a, 2b and 2c reveals the relatively finer macrostructure in the case of electron beam welding , but its microstructure is almost entirely



a) Electron beam weld

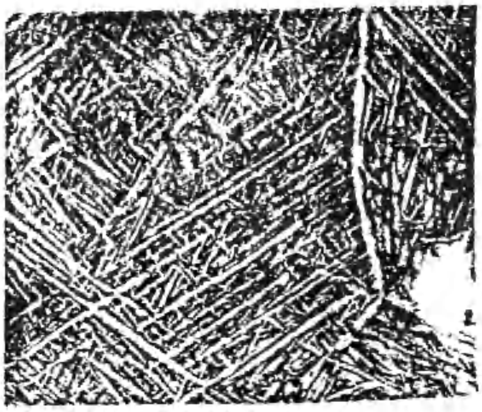


b) Automatic GTA weld



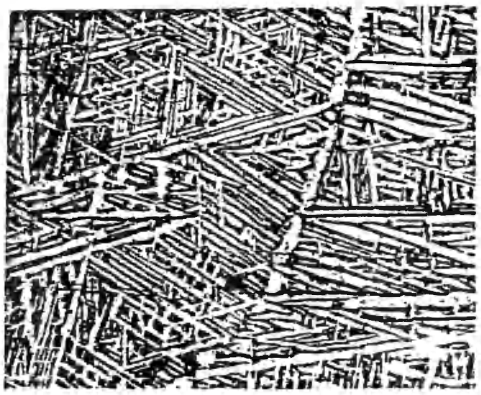
c) Manual GTA weld

Fig 5 : Microstructures of the fusion zone after post-weld heat treatment at 700° C (500 X)



a) Electron beam weld

b) Automatic GTA weld



c) Manual GTA weld

Fig 6 : Microstructures of the fusion zone after post-weld heat treatment at 900° C (500 X) martensitic. It must, however, be remembered that, unlike Fe-C alloys, titanium-martensites are not very hard and brittle. The combined effect of these factors results in marginally higher as-welded ultimate and yield strengths and tensile elongation for the electron beam welds.

The hardness distributions across the different weldments are plotted in Figs. 7, 8 and 9. In agreement with the generally refuced weld metal ductility, the hardness curves in the as-welded condition also show a distinct hardening effect in the weld and near heat-affected zones. This is particularly true in the case of electron beam welds.

The benefits of post-weld heat treatment are not quite realised at 700°C. In fact Table 2 shows that the treatment at 700°C has led to a reduction in ductility from the as-welded condition. The yield and ultimate tensile strengths are also reduced but only very slightly. Similarly from Figs. 7,8 and 9 it can be seen that the FZ hardness has decreased on heat treating at 700°C but not appreciably. It is only when the heat treatment is carried out at 900°C that tensile elongations close to that of the base metal are obtained. This accompanied by strength reduction to values somewhat lower than in the base metal. it is also worth noting that the treatment at 900°C has nearly equalised the FZ properties in weldments produced by all three processes.

These results correlate well with the microstructural observations made earlier. Although evidence dying of the process is not easily observable in the microstructure after heat treatment at 700°C (Fig. 5), its occurance is demonstrated by the trend shown in the mechanical properties as they change during the heat treatment. The hardness curves (Figs. 7,9) show that after heat treatment at 700° C the hardening due to aging is compensated at softening due to Coarsening effects though the latter are not as yet prominent; In other words overaging effects just begin to be apparent at the heat treatment temperature of 700°C. The fact that the structure still remains predominantly acicular with a high aspect ratio is responsible for the ductility continuing to be low.

Further overaging occurs as the alpha plates thicken during the 900°C treatment leading to a distinct drop in hardness as seen in (Figs. 7.9). This is accompanied by an increase in ductility and a reduction in strength.

The mechanical behaviour of the high-temperature heat-treated weldments can be explained in terms of the grain-boundary alpha phase present in conjunction with a coarse transformed beta intragranular microstructure, Fig. 6. Precipitation of the alpha phase at the prior-beta boundaries could not be observed in the specimens heat treated at 700°C. However, after treatment at 900°C a relatively thick, though discontinuous alpha layer develops at the grain boundaries. The intragranular structure also consists of coarse alpha plates in a beta matrix and it is known that little strength difference exists between such a coarse transformed-beta structure and the grain boundary alpha. In such a condition it is reasonable to expect that slip initiating in the relatively weaker grain boundary alpha dur-

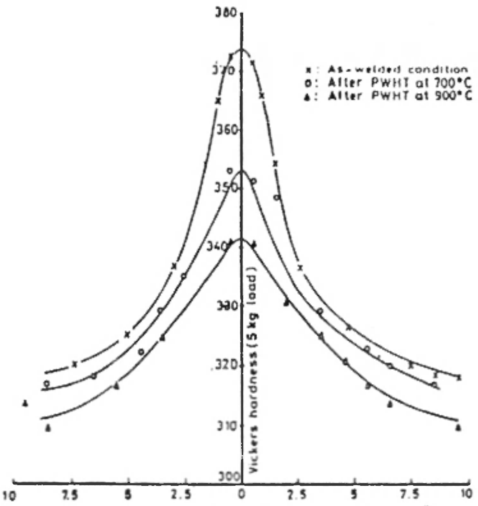


Fig 7 : Hardness distribution Across elcetron beam weld

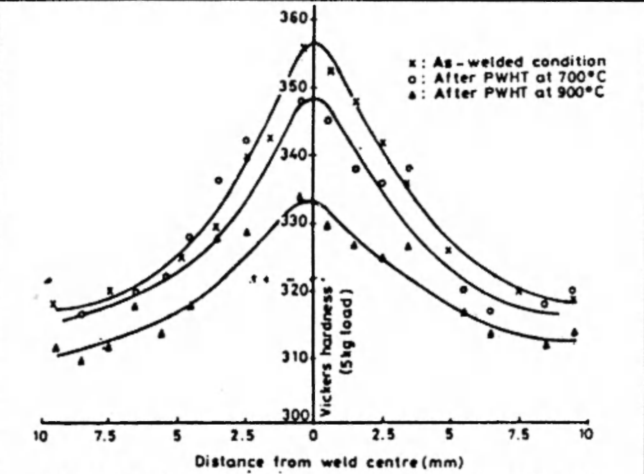


Fig 8 : Hardness distribution Across Automatic GTA weld

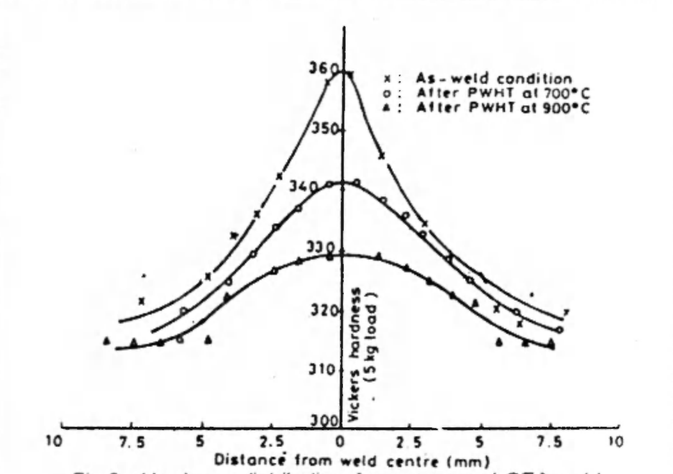


Fig 9 : Hardness distribution Across manual GTA weld

ing tensile loading can be easily accomodated intragranularly. This prevents slip concentration at the grain boundary interfaces and improves macroscopic weld ductility at reduced strength levels (6).

In general, however, it must be noted that lamellar alpha microstructures (as in as-welded and heat-treated weldments) posses reduced ductility in relation to structures containing equiaxed alpha (as in the base metal, Fig. 1). This is because, in titanium alloys, on account of their low work-hardening rate, the growth of microvoids during tensile fracture is quite rapid; the primary component of the fracture strain is thus the strain to nucleate microvoids. Since voids are known to nucleate on alpha-beta interfaces, a high aspect ratio of the alpha phase increases the interfacial area per unit volume and reduces the tensile fracture strain (8). This is the reason why, even in the high temperature heat-treated weldments, the tensile elongation is still not as high as that exhibited by the base metal.

The similarity in mechanical properties in weldments made by all three processes after post-weld heat treatement at 900°C is not surprising because, as mentioned earlier, the microstructures in Fig. 6 also appear identical. Clearly, when the weldments are held at a temperature of 900°C, a near-equilibrium condition is reached for a heat treatment period of three hours.

CONCLUSIONS

(1) The alpha-beta Ti-6A1-4V alloy suffers a drop in fusion zone ductiletey on account of beta grain growth and martensite formation during welding.

(2) Despite a relatively smaller prior-beta grain size, electron beam weldments show only a marginally higher tensile elongation.

(3) Post-weld heat treatment at a temperature high in the alpha-beta phase field is required for restoring adequate ductility.

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