FRACTURE BEHAVIOUR OF Ti-6Al-4V WELDMENTS AND ITS RELATION TO MICROSTRUCTURAL FEATURES

by

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(This paper, presented at the National Welding Seminar-1995 at Cochin, has won the "Mrs. D.M. Panthaki Award for welding of non-ferrous metals")

THEME

Tensile fracture of α–β titanium alloy weldments is sensitive to microstructure. Thus, though it occurs macroscopically at low ductility, it follows a void coalescence mechanism. The paper describes work in which the alloy Ti-6Al-4V has been welded and heat-treated under different conditions. The role of the morphology and distribution of the α and β phases on the mode of tensile fracture has been discussed. Fracture toughness measurements of the fusion zone have shown that (i) the as-welded structure resists crack growth better than the base material; and (ii) post-weld treatment high in the α–β field (say at 900°C) confers a further improvement over the as-welded level, while treatment at intermediate temperatures like 700°C could actually be detrimental. The paper explains the reasons for this behaviour in terms of microstructural features and scanning-electron fractographic observations.

INTRODUCTION

It has been known for long that welding of α–β titanium alloys results in a reduction in ductility even though there is no adverse effect on strength which, in fact, may increase(1). This is generally attributed to a large prior-β grain size and an acicular martensitic microstructure (2). It is therefore necessary to resort to a post-weld heat treatment to impart sufficient ductility to the fusion zone.

The strengthening of the microstructure and the low level of ductility had led to an earlier belief that tensile fracture of these weldments under monotonic loading conditions occurs by cleavage (3,4), though this fracture mode is not commonly observed in α–β titanium alloys. Subsequently, however, observation at higher magnification has revealed that dimples could be noticed, evidencing microscopically ductile fracture occurring by a void coalescence mechanism (5).

The fracture behaviour associated with the propagation of a pre-existing crack has also been studied widely in α–β titanium alloys. Previous studies (6, 7, 8) have demonstrated the superiority of microstructures containing a large percentage of acicular α, which has been attributed, for transgranular fracture, to the extended α–β interfacial area available for crack growth in lamellar structures (9) and, for intergranular fracture, to the increased fracture path along prior-β grain boundaries (10). The role of grain boundary α in intercrystalline fracture has also been described (11).

Although fracture toughness has been widely studied as a function of thermal and mechanical processing conditions, there have been relatively few investigations in respect of welded joints. Further, there has been no clear correlation of toughness with any microstructural feature of the weldment.

The current paper reports work in which the α–β titanium alloy Ti-6Al-4V was welded using different heat inputs and subsequently heat treated at different temperatures. Both tensile fracture behaviour and fracture toughness of the weldments in the various welded and heat-treated conditions have been studied in detail. Basically, the paper correlates the two types of fracture behaviour with microstructural fractures.

EXPERIMENTAL DETAILS

The work was carried out on a 6 mm thick α–β processed Ti-6Al-4V sheet containing Al-6.01 and V-4.01 wt%.

Three welding processes were used : (1) manual gas tungsten - arc welding (M-GTAW) in a 1.5 meter diameter hemispherical glove-box filled with argon; (2) automatic gas tungsten-arc welding (A-GTAW) with auxiliary trailing and backing gas shields; (3) electron beam welding (EBW). The heat inputs were, respectively, 670, 480 and 315 J/mm. Autogenous full-penetration bead-on-plate welds were made in all cases. Post-weld heat treatment (PWHT) was performed at 900°C and at 700°C for 3 hours, in a vacuum of 10^{-5} torr.

The structural changes were followed through light microscopy and transmission electron microscopy (TEM), using standard specimen preparation procedures. Longitudinal all-weld specimens were used for tensile testing. The tensile fracture faces were observed in a scanning electron microscope (SEM).

As a measure of fracture toughness, the elastic-plastic parameter J_{IC} was used. The tests were performed in accordance with ASTM E813-89 (12). Single-edge notched compact tension specimens with the geometry shown in **Fig.1** were used. Fatigue pre-cracking and subsequent loading for crack extension were carried out in a 100 kN servo-hydraulic

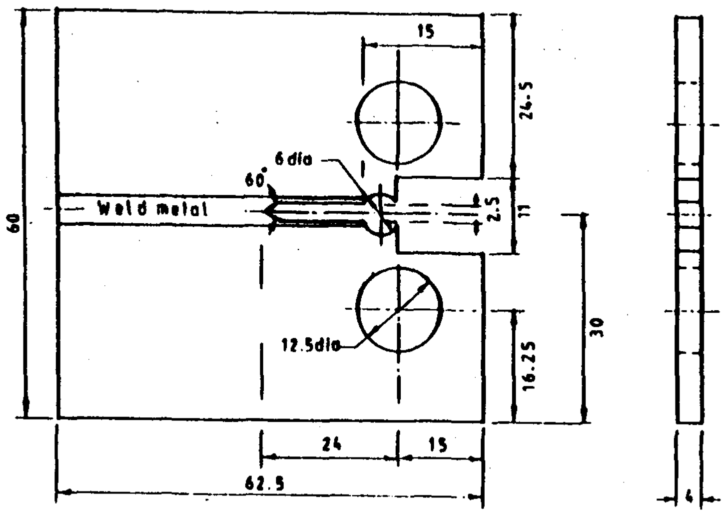


Fig. 1 : Compact tension specimen

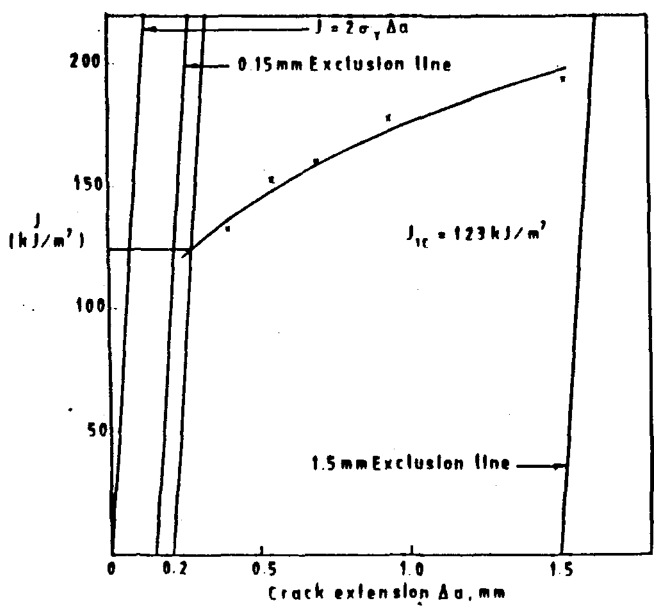


Fig. 2 : J_R curve for Ti-6Al-4V manual GTA weld, post-weld heat treated at 900°C

testing machine. A plot of J - integral vs crack extension Δa is shown in **Fig. 2** and illustrates how J_{IC} is determined. After J_{IC} testing the fracture faces were examined in an SEM.

RESULTS AND DISCUSSION

Microstructures

The base metal microstructure shows a *mixture* of primary α grains in a transformed-β matrix (**Fig. 3**). In the as-welded condition, the fusion zone (FZ) exhibited fine acicular microstructures

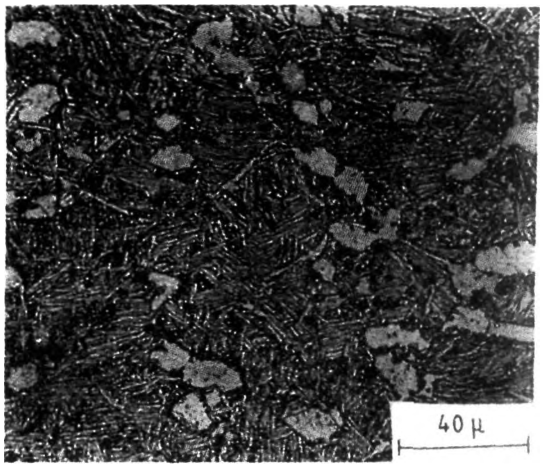
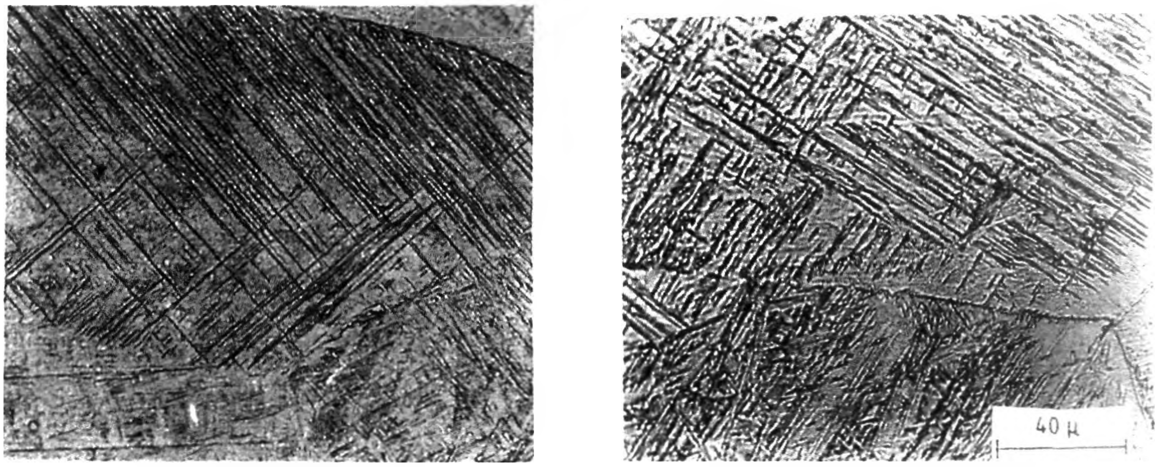
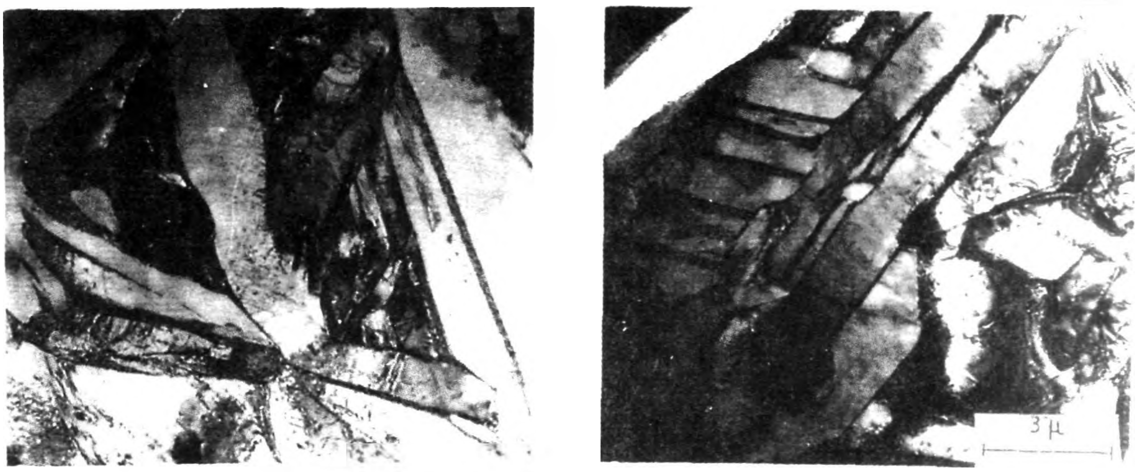


Fig. 3 : Light micrograph of the base material



(a) : Electron beam weld **(b)** : Manual GTA weld

Fig. 4 : Light micrographs of fusion zone, as-welded condition



(a) : Electron beam weld **(b)** : Manual GTA weld

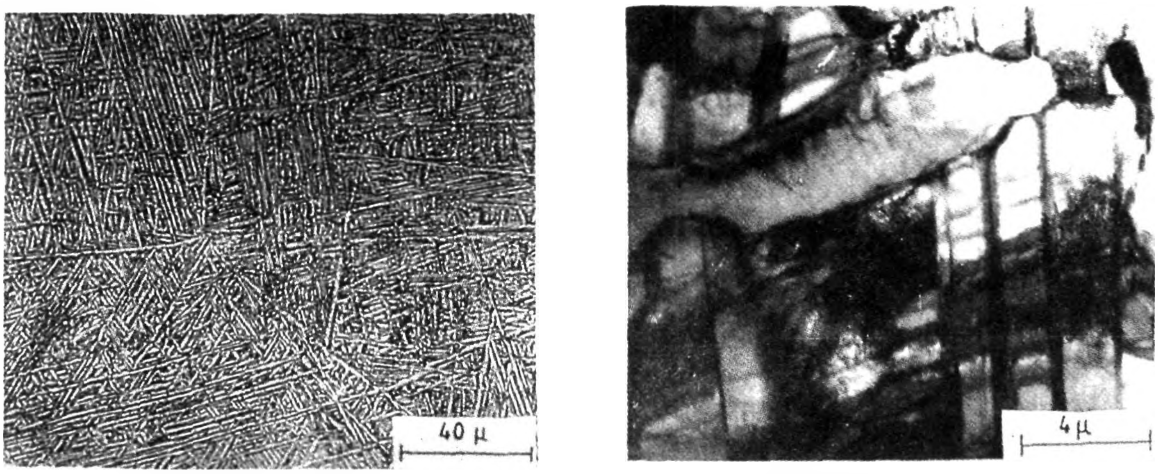
Fig. 5 : Electron micrographs of fusion zone, as-welded condition

for all three welding conditions, see **Fig. 4a & 4b**. The relatively slower cooling in M-GTAW in comparison to EBW is reflected in the greater intragranular coarsening in the M-GTA weld. The difference is also revealed in the transmission electron micrographs (TEM's) shown in **Fig. 5a & 5b**. The former is almost entirely martensitic, while the slower-cooled GTA weld exhibits a mixture of martensite and a colony structure consisting of α plates separated by thin strips of retained β, the latter confirmed by selected area diffraction. The A-GTA weld FZ showed features intermediate between those of EB & M-GTA welds. The microstructures after PWHT at 700°C and 900°C are shown in **Fig. 6a & 6b** and **Fig. 7a & 7b**, respectively, for M-GTA welds. Electron beam and A-GTA welds showed PWHT microstructures similar to these: a Widmanstatten α–β structure within the grains and a discontinuous α film at the grain boundaries. However, the coarsening effect was much less pronounced at 700°C than at 900°C, compare **Fig. 6a** with **7a**, and **6b** with **7b**. Some plates of martensite may still be observed in **Fig. 6b**, while in contrast the α plates in **Fig. 7b** are completely free of dislocations.

Mechanical Properties

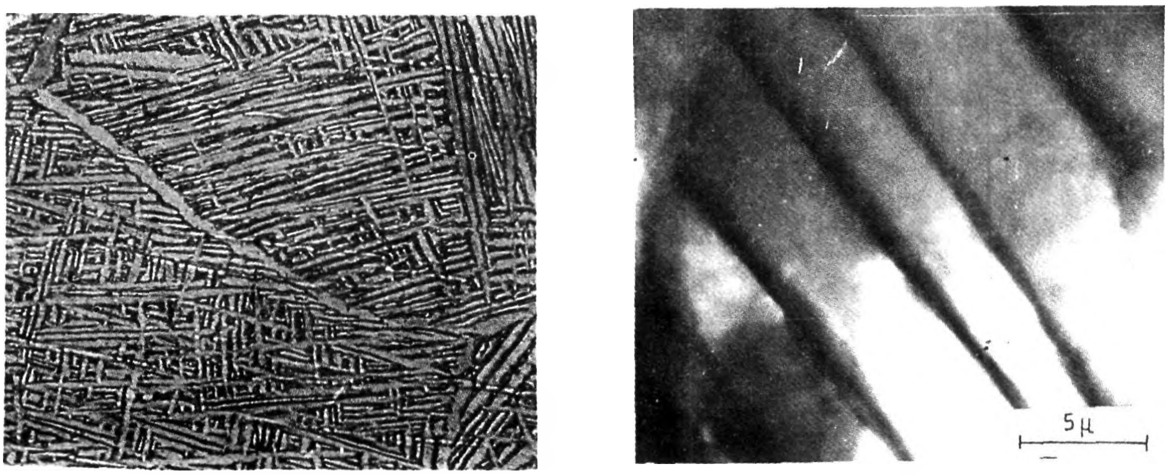
Tensile Properties

The tensile test results are listed



(a) : Light micrograph **(b)** : Electron micrograph

Fig. 6 : Microstructure of fusion zone after PWHT at 700°C



(a) : Light micrograph **(b)** : Electron micrograph

Fig. 7 : Microstructure of fusion zone after PWHT at 900°C

in **Table I**. The most significant feature is the reduction in ductility as a result of welding, which is to be attributed to the large prior-β grain size and the acicular microstructure. Note that PWHT at 700°C is not beneficial and, in fact, leads to a reduction in tensile elongation from that in the as-welded condition. Only a heat treatment at 900°C is able to raise the ductility to a value comparable to that of the base material. The poor ductility after the 700°C treatment is probably a result of the fact that the microstructure has not sufficiently coarsened at this temperature and continues to exhibit a high aspect ratio of the α plates. It is known that under such conditions tensile fracture strain is low since a greater α–β interfacial area per unit volume is available for void nucleation (7).

TABLE I - Fusion Zone Properties of Ti-6Al-4V Weldments

Welding Process	Condition	YS (MPa)	UTS (MPa)	Elong (%)	J_{IC} (kJ/m^2)
EBW	As-welded	894	1005	9.5	79
	PWHT at 700°C	884	997	8	67
	PWHT at 900°C	805	936	12	126
A-GTAW	As-welded	880	975	8	98
	PWHT at 700°C	–	–	–	54
	PWHT at 900°C	784	896	12.5	124
M-GTAW	As-welded	873	986	8.5	97
	PWHT at 700°C	866	965	6.5	55
	PWHT at 900°C	800	925	12	123

Note : 1. PWHT stands for post-weld heat treatment.
2. All tensile results are an average of 2 tests each.
3. For comparison, the base metal properties are : 857 MPa (YS), 973 MPa (UTS), 14.5% (Elong.) and 74 kJ/m^2 (J_{IC})

Tensile Fracture Behaviour

The scanning electron micrographs (SEM's) of the tensile fracture faces are given in **Fig. 8 & 9**. All as-welded tensile specimens from the FZ fractured in transgranular fashion, **Fig. 8a - c**. The fractures are irregular and show some faceting as has been observed in other α–β titanium alloys (5,13). However, increased magnification reveals the presence of predominantly equiaxed dimples of varying sizes.

PWHT at 700°C produces little change in fracture appearance, **Fig. 9a**. The surface appears mainly transgranular; while evidence of faceting still remains, higher magnification shows the existence of dimples. On the other hand, PWHT at 900°C reveals a larger amount of intergranular fracture and no faceting can be observed, **Fig. 9b**

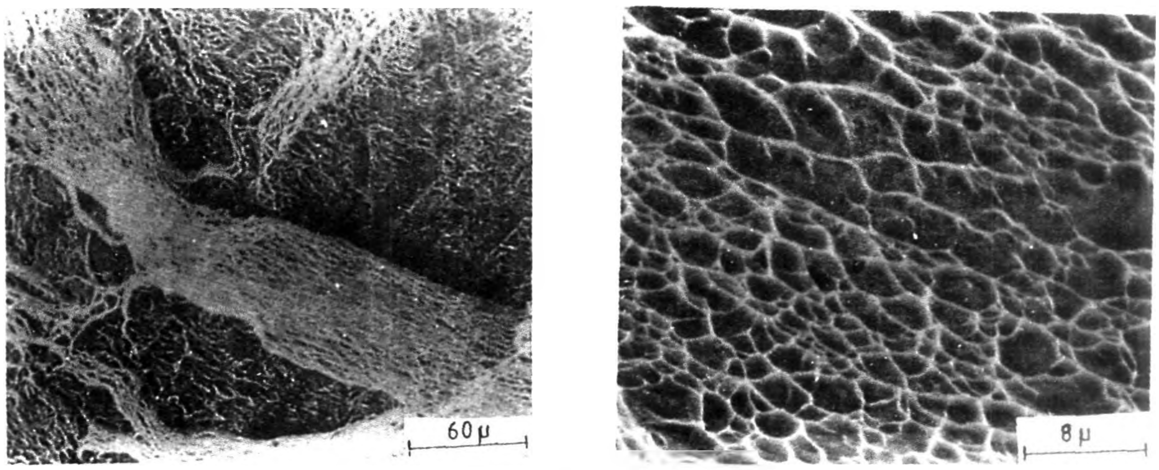
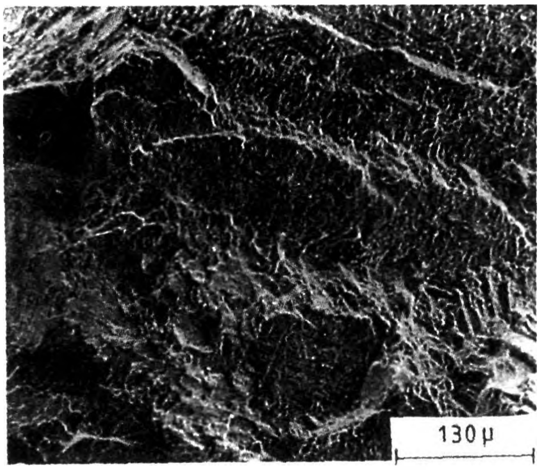
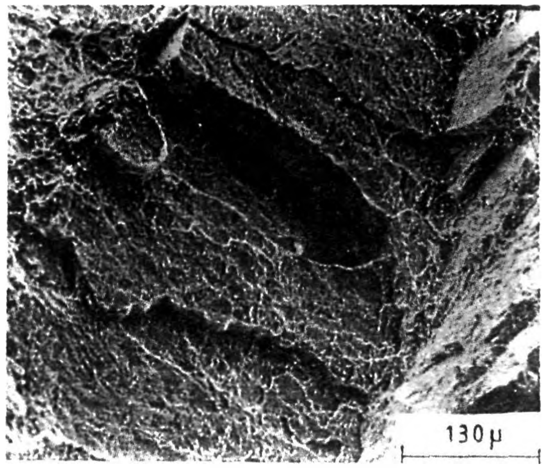


Fig. 8a : Electron beam weld

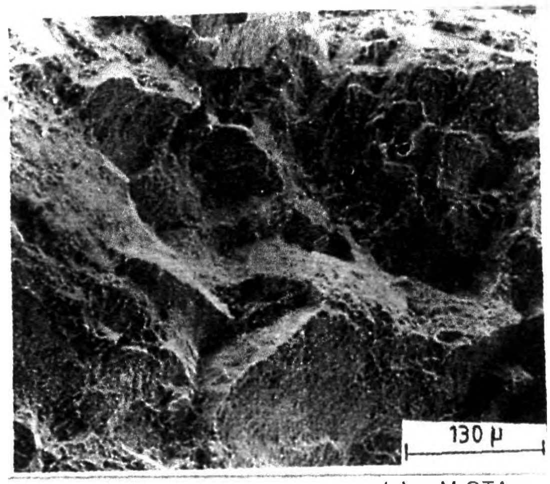


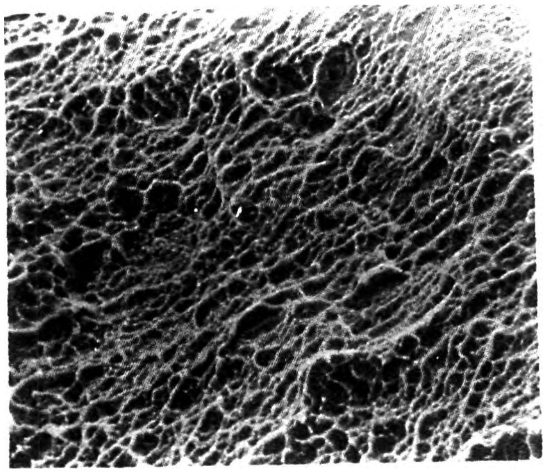
(8b) : Automatic GTA weld



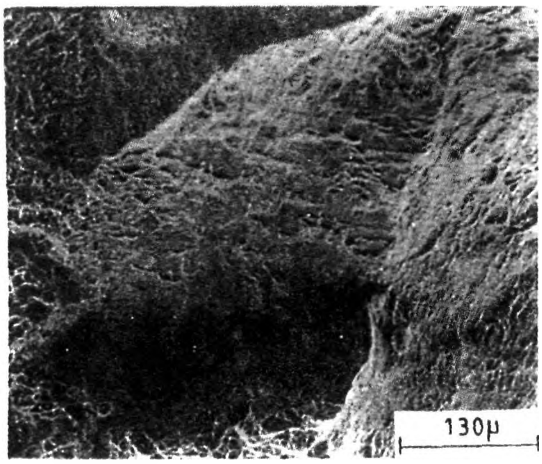
(8c) : Manual GTA weld

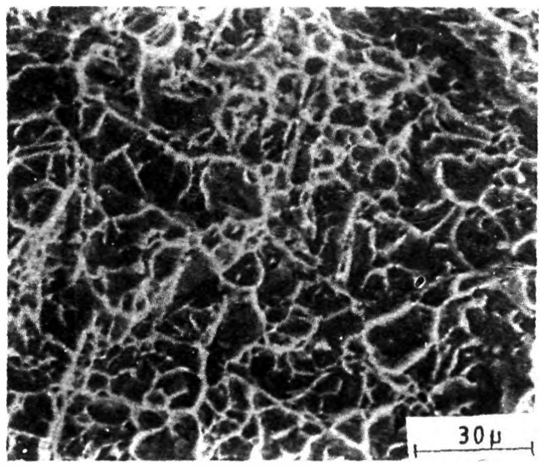
Fig. 8 : SEM fractographs of Ti-6Al-4V fusion zone tensile samples in as-welded condition





(a) : M-GTA weld after PWHT at 700°C





(b) : M-GTA weld after PWHT at 900°C

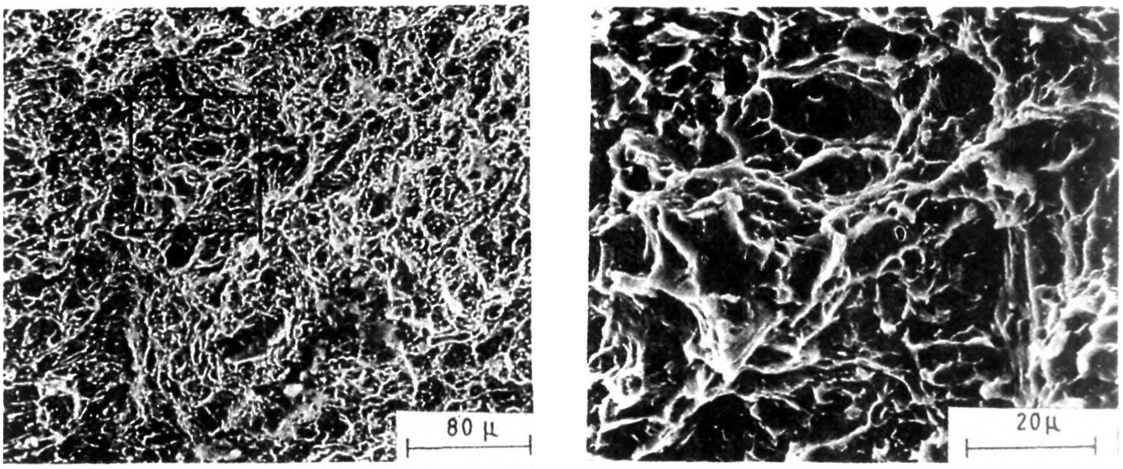
Fig. 9 : SEM fractographs of Ti-6Al-4V fusion zone tensile samples

Dimples are seen again at higher magnification.

The tendency to partial intergranular fracture after the 900°C heat treatment (**Fig. 9b**) may be attributed to the presence of the thick, though discontinuous, grain boundary α phase in the microstructure. However, this does not result in a reduction of ductility; in fact, it is improved after the 900°C treatment. This is presumably a result of the presence of the coarse intragranular transformed β microstructure. Little strength difference exists between such a coarse transformed–β structure and the grain boundary α. It is reasonable to expect that slip initiating in the weaker grain boundary α during tensile loading can be easily accommodated intragranularly. This prevents slip concentration at the grain boundary and improves macroscopic weld ductility at reduced strength levels (1). This is consistent with the observation that a good proportion of the fracture in the 900°C heat-treated condition is transgranular.

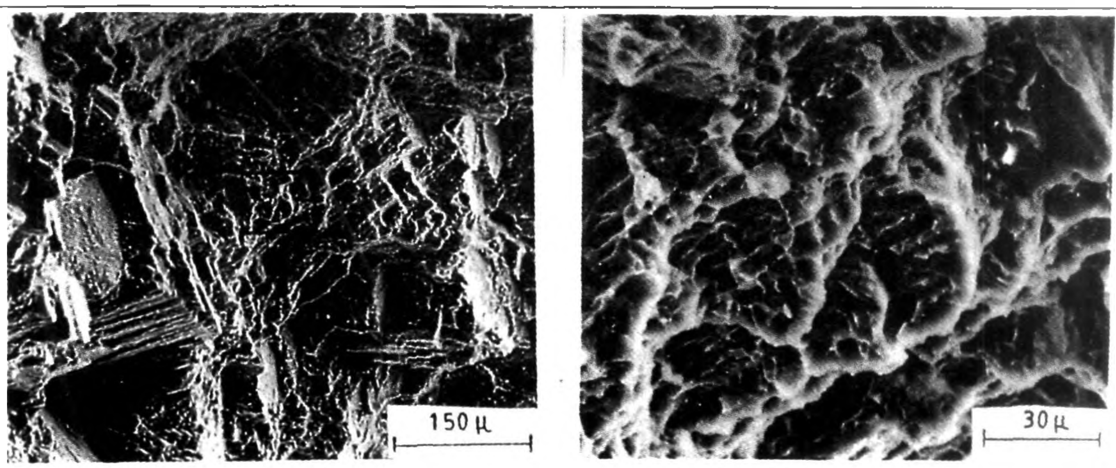
Fracture Toughness

The fracture toughness (FT) results are given in Table I. The corresponding SEM's of the fracture faces are shown in **Fig. 10-13**. It is apparent that the as-welded condition in all three cases is characterised by a higher FT than that of the base material. While PWHT at 900°C has resulted in a further increase



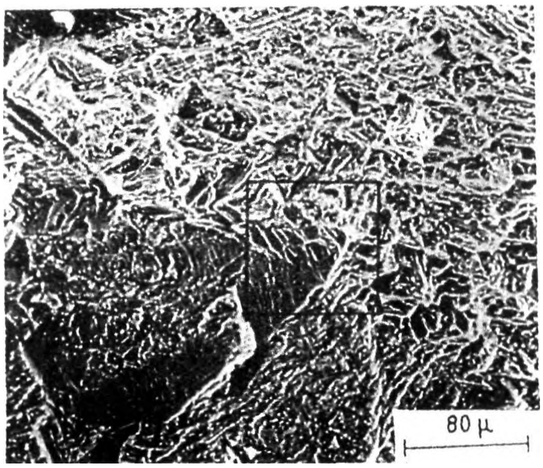
(a) (b)

Fig. 10 : SEM fractographs of Ti-6Al-4V base metal FT samples

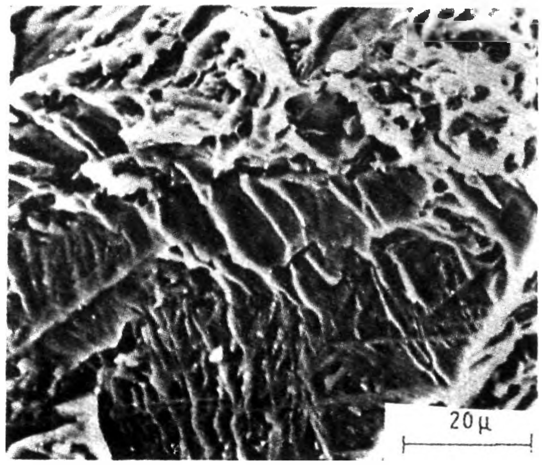


(a) (b)

Fig. 11 : SEM fractographs of fusion zone FT sample of EB weld, as-welded condition [Box in (a) magnified in (b)]



(a)



(b)

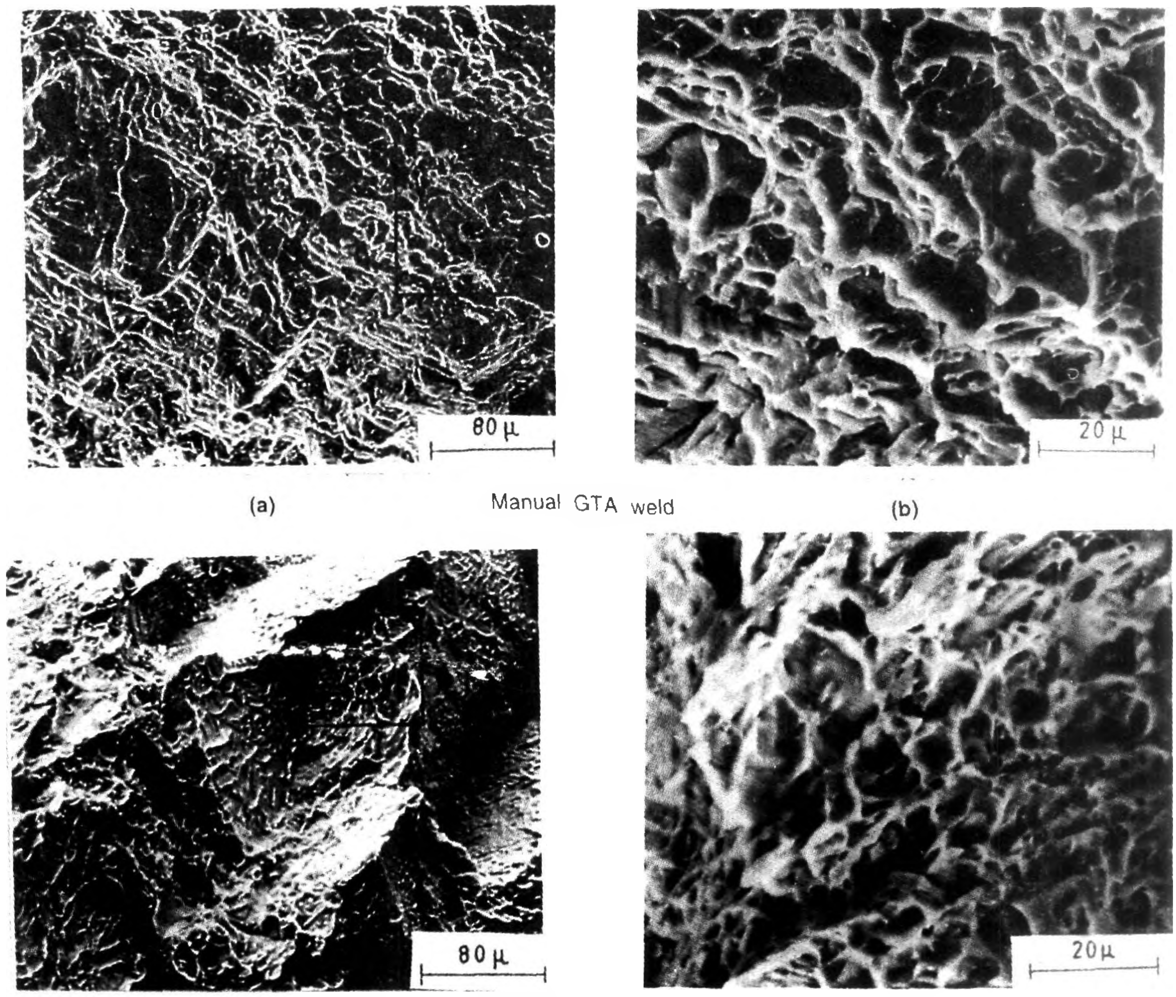
Fig. 12 : SEM fractographs of fusion zone FT sample of EB weld, after PWHT at 700°C

in FT, the treatment at 700°C has led to a substantial reduction in energy absorbed.

The higher FT of the as-welded FZ in relation to the parent material is, no doubt, due to the fully acicular microstructure. On the other hand, the base metal microstructure consists of equiaxed α and transformed-β regions (**Fig. 3**). The relative superiority of plate-like microstructures, both for transgranular and intergranular fracture, is well known. The fractographs of the base metal, **Fig. 10a & b**, show a relatively flat fracture face with only little deviation in crack path. In contrast, the fracture faces of the as-welded FZ, in **Fig. 11a & b**, exhibit greater diversions in crack path, and many features of ductile fracture like tear ridges (**Fig. 11a**) and dimples surrounded by tear ridges (**Fig. 11b**). The rougher topography of **Fig. 11a** is consistent with the higher FT of the weld metal.

On PWHT at 700°C, the FT suffers a drastic reduction from the as-welded value. The SEM's of the fracture face (**Fig. 12**) show a mixture of transgranular and intergranular fracture. Microstructurally, two features are important, as seen in **Fig. 6a** : the intragranular coarsening which at this temperature is not pronounced and the grain boundary α layer which also has not developed to any significant thickness. The latter is likely to promote intergranular fracture, as the grain boundaries are known to offer ideal sites for void formation (14). The fact that the interfacial α phase, though softer than the transformed/aged β structure, is very thin means that it will be constrained by the surrounding material and cannot deform freely. The energy required for intergranular crack propagation is thus effectively reduced. Considering transgranular fracture, the poor energy absorption is believed to be related to the fineness of the matrix microstructure. While the generally superior toughness of lamellar structures has never been in doubt, it has often been suggested that such structures should be sufficiently coarse in order to be effective in raising FT. The plates must be thick enough to turn a propagating crack but short enough and close enough together to cause frequent changes in crack growth direction (8). It is surmised that, after PWHT at 700°C, the plate thickness has developed to such a small extent that it is not able to divert a propagating crack. In the SEM's in **Fig. 12**, there is little evidence of plastic flow or of crack deviation.

The highest FT in the current investigation was obtained from the PWHT at 900°C, Table I. The fracture surfaces, **Fig. 13a-d,** exhibit a mixture of well-defined intergranular rupture along prior-β grain boundaries (**Fig. 13c**) along with isolated regions of



(a) Manual GTA weld (b)

(c) Automatic GTA weld (d)

Fig. 13 : SEM fractographs of fusion zone FT sample after PWHT at 900°C [Box in (a) & (c) magnified in (b) & (d) respectively]

transgranular fracture (**Fig. 13a**). The ductile nature of the fracture is clearly visible in the regions of tearing and void formation, in **Fig. 13b**. The intergranular part of the fracture is also ductile as revealed by the presence of microvoids in the higher magnification picture (**Fig. 13d**).

The fact that fracture toughness after PWHT at 900°C is approximately double that obtained after the treatment at 700°C can be traced to the difference in the two microstructures, **Fig. 6a & b** and **Fig. 7a & b.** The difference in coarsening between the two structures, both for the intragranular α plates and for the grain boundary α layer, is unmistakable.

In the transgranular part of the fracture, the thicker plates in the basketweave structure resulting from the 900°C treatment can cause deviations in crack path more effectively and more frequently than the thinner platelets developed at 700°C. Considering the intergranular regions of the fracture, the thicker grain boundary α layer generated at 900°C may be expected to undergo a greater degree of plastic flow on account of the reduced constraint imposed by the neighbouring intragranular structure. This is consistent with previous studies on the role of the intergranular α, which showed that thicker grain boundary α layers increase the energy needed for fracture (11).

CONCLUSIONS

1. Welding of the α-β Ti-6Al-4V alloy reduces its ductility which can be adequately improved only by PWHT at about 900°C.
2. Though tensile fractures show cleavage features such as facets, they occur in fact by a microscopically ductile, void coalescence mechanism as revealed by dimples at higher magnification.
3. The as-welded condition is characterised by a fracture toughness higher than that of the base material.
4. PWHT can further improve FT, but only when the temperature of treatment is high in the α–β region close to the β-transus.
5. The lower-temperature heat treatment, for example at 700°C, may actually prove to be detrimental by reducing toughness and ductility from the as-welded levels.

ACKNOWLEDGMENT

The above work was part of a larger project sponsored by the DST, whose financial support is gratefully acknowledged.

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