EFFECT OF STRUCTURAL CHANGES ON TENSILE PROPERTIES IN AN ELECTRON BEAM WELDED AND HEAT-TREATED Ti-Al-Mn ALLOY

By

K. Keshava Murthy and S. Sundaresan

Department of Metallurgical Engineering, Indian Institute of Technology, Madras

(Presented during NWS'94 at Jamshedpur)

INTRODUCTION

The high strength-to-weight ratio of titanium alloys has contributed to their extensive use in the aerospace sector. More recently, however, their excellent resistance to corrosion and generally superior mechanical properties have resulted in their increasing use in the chemical industry as well as by the medical profession. Many of these applications involve welded joints and from this point of view the alpha-beta titanium alloys are considerably more difficult to join than the alpha or the beta alloys. Within the alpha-beta family, though many highstrength alloys have been developed to have greater fracture toughness than the standard alloy Ti-6Al-4V, their use has been limited because of poor weldability [1]. This is because of invariably low ductility in the as-welded condition and there is some evidence that, with the addition of increasing amounts of beta stabilizers, the ductility deteriorates further [2]. This has led to the design philosophy in the erstwhile Soviet Union involving the development of titanium alloys with moderate strength but appreciably greater ductility for welded fabrications subjected to complex stresses. While heavier cross-sections can be used for maintaining load-carrying capacity, the ductility would ensure satisfactory weldability [3].

One of the medium-strength alloy in the above category is OT4-1 [Ti-2.5Al-1.8Mn nominal composition]. Though manganese is a strong beta stabilizer, the alloy content is low. Russian literature classifies it as a pseudo-alpha alloy [4], which can be used upto 350°C in the welded condition.

The weldability of the OT4-1 alloy has been demonstrated both in electron beam [5] and gas tungsten-arc welding [6]. However, as in other alpha-beta titanium alloys, while the strength of the weld metal matches, or even exceeds, that of the parent metal, there is a significant reduction in ductility in the fusion zone [6] necessitating a post-weld heat treatment. With regard to the temperature of such heat treatment, however, the evidence available in literature is somewhat confusing. For example, it has been shown that post-weld heat treatment at 450-500°C is superior to one at 650°C from the point of view of properties such as tensile strength, fracture toughness and fatigue strength [7]. It has also been reported that if the welded joints are quenched from temperatures in the range of 800 to 950°C, a strengthening effect results from subsequent annealing at 450°C [8]. Evidence regarding the fusion zone ductility as a consequence of post-weld heat treatment is significantly lacking. There is also no information concerning again at higher temperatures, as has been recommended, for expample, for the higher-strength alpha-beta titanium alloy weldments [9,10].

The current paper describes fundamental work carried out on the medium-strength OT4-1 titanium alloy which, after electron beam welding, was subjected to different heat treatments. The objective was to optimise welding and heat treatment procedures and to correlate systematically the microstructures in the various conditions with the corresponding mechanical properties.

Experimental Details

The titanium alloy used in the investigation was a 10mm thick plate whose chemical composition and tensile properties are given in **Table 1 and 2** respectivaly. Autogenous full-penetration bead-on-plate welds were deposited by electron beam welding using the parameters listed in **Table 3.** Just prior to welding the base metal coupons were pickled in a 25 HNO_3 + 10HF +65H_2O solution

Table 1 : Chemical composition of the alloy OT4-1

Al	Mn	C	Fe	Si	Zr	H	O	N	Ti
2.34	1.45	<0.01	<0.1	<0.01	<0.1	0.003	0.09	<0.015	Balance

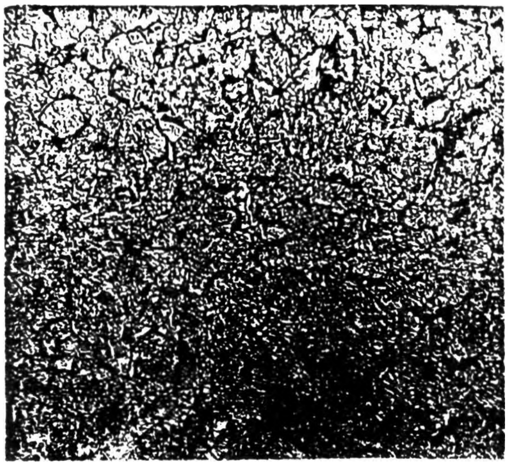

Fig. 1 : Microstucture of the base metal (500X)

Fig. 2 : Microstucture of the fusion zone (50X)

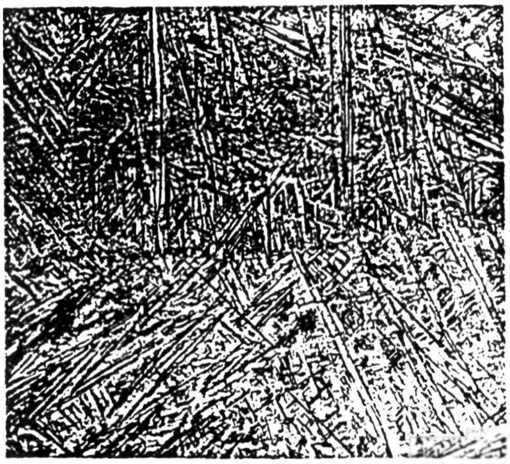

Fig. 3 : Microstucture of the fusion zone, as-welded condition (500X)

Fig. 4 : Microstucture of the fusion zone after post-weld heat treatment at 650°C (500X)

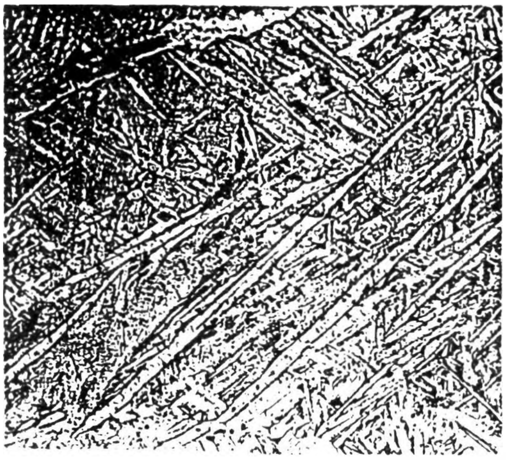

Fig. 5 : Microstucture of the fusion zone after post-weld heat treatment at 850°C (500X)

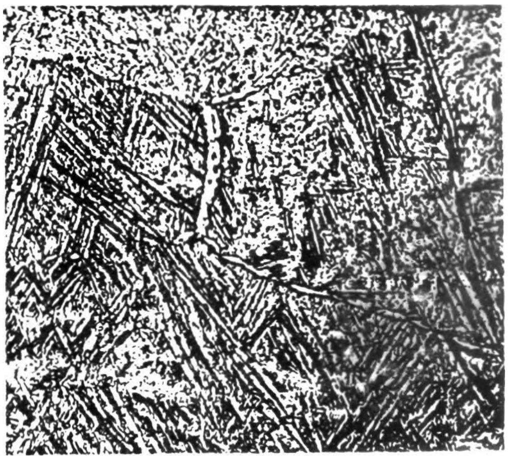

Fig. 6 : Microstucture of the fusion zone after duplex post-weld heat treatment at 850+540°C (500X)

Table 2 : Tensile Properties of the base material

No.	Condition	Yield strength (MPa)	UTS (MPa)	Elongation %
1	Hot rolled (as received)	604	728	9.5
2	After annealing 650°C/3h/FC	532	648	14
3	After Annealing 850°C/3h/FC (FC = Furnace cool)	453	570	15.5

and degreased with acetone. The welded strips were subjected to post-weld vacuum heat treatment at different temperatures, the details of which are given in **Table 4.** Microstructural characterisation was performed using light microscopy. The etchant used was a 2:1 HF-HNO_3 aqueous solution. Mechanical testing of the welds included hardness survey and tensile testing. Vicker's hardness was measured with a 5 kg load at 0.5mm intervals across the welds. For tensile testing longitudinal all-weld cylindrical specimens were cut from the weld metals and machined to a diameter of 3mm in the gauge length portion. Macroetching showed that the weld metal occupied a little over 80% of the gauge length cross-section. Tensile testing was conducted in a 100 kN Instron machine at a cross-head speed of 0.05 mm/min. The tensile fracture faces were examined in a scanning electron microscope for obtaining possible information regarding the nature of the fracture processes in different conditions.

RESULTS AND DISCUSSION

Structural Features

The microstructure of the base material is given in **Fig. 1**. It shows recrystallized, equiaxed primary alpha grains with a small amount of beta phase at the grain boundaries. It must be noted that the concentration of beta stabilizing element, viz., manganese, is quite low in the alloy.

During welding, the material in the heat-affected zone is raised to various high temperatures above and below the (alpha+beta)/beta transus temperature which for this alloy is 920°C. In the fusion zone, solidification occurs as beta crystals, which on further cooling decompose either martensitically to the alpha-prime phase or by diffusion processes to the alpha phase. The macrostructure of the weld metal, given in **Fig. 2**, shows columnar beta grains that have grown from the heat-affected zone. The weld metal microstructure, **Fig. 3**, reveals a fine acicular morphology that appears to be largely, if perhaps not fully, martensitic. The small heat input and rapid heating characteristic of electron beam welding result in a cooling rate that would provide little time for nucleation and growth processes required for the beta to alpha transformation. Further evidence is provided by the absence of any alpha allotriomorphs along the grain boundry running midway across the micrograph in Fig. 3. One feature of the as-welded structure not distinctly revealed in the micrograph in Fig. 3 is the presence of a small amount of retained beta phase delineating the martensite plates. The presence of the beta phase was confirmed separately by transmission electron microscopy using selected area diffraction.

The microstructure after post-weld heat treatment at 650°C, given in **Fig. 4**, shows practically no evidence of intragranular coarsening nor the development of any grain boundary alpha phase. However, when the heat treatment temperature was raised to 850°C, the structure be-

Table 3 : Welding parameters for electron beam welding of OT4-1 alloy

Beam voltage	:	130 kV
Beam current	:	68.5 mA
Travel speed	:	35 mm/sec
Heat input	:	254 J/mm
Vacuum	:	10^{-5} torr

Table 4 : Post-weld heat treatment parameters for OT4-1 welds

1. 650°C /3h/FC
2. 850°C/3h/FC
3. 850°C/1h/WQ +540°C/16h/FC

(FC = Furnace cool, WQ = water quench)

came considerably coarsend, **Fig. 5**. The thick Widmanstatten alpha plates, however, do not exhibit the "basket weave" arrangement typical of alpha-beta titanium alloys with a higher alloy content such as Ti-6Al-4V. It is important to note the layer of alpha phase that has formed along the prior beta grain boundary at the top left of Fig.5. It must, however, be mentioned that the alpha phase along the grain boundary even after the 850°C heat treatment was not continuous in the rest of the microstructure. The structure after a duplex post-weld heat treatment consisting of solution treatment at 850°C followed by again at 540°C is shown in **Fig. 6.** While the coarsening is considerably greater than after the treatment at 650°C, it has occurred to a lesser degree than in the specimen aged at 850°C for 3 hours (Fig. 5). it may also be observed that the grain boundary alpha film is thinner than in Fig.5 and is discontinuous. There are some regions of the grain boundary which show no alpha precipitation. Another interesting feature in Fig. 6 is that the alpha plates are more clearly delineated by the surrounding beta films on both sides than in Fig. 5

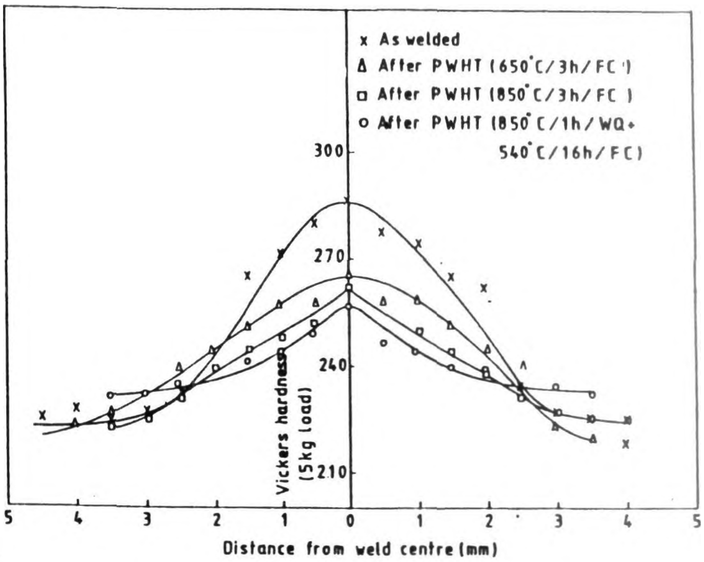

Fig. 7 : Hardness distributions across the weld

Mechanical Properties

The tensile properties of the base material listed in Table 2 show the high structure sensitivity of the alloy.

In the alpha-beta hot-rolled condition, in which it was received, it has an ultimate tensile strength of 728 MPa and an elongation of 9.5%. Annealing at 650°C, the recommended annealing temperature for OT4-1 welded joints [11], reduces the tensile strength to 648 MPa but increases the ductility to 14%. These trends continue when the annealing temperature is raised to 850°C, which lies high in the alpha-beta phase field.

The tensile properties of the welded joints are given in **Table 5** and the hardness distributions across the weld in **Fig. 7**. In the as-welded condition, there is a steep increase in yield and tensile strengths, while there is an equally significant reduction in ductility in relation to the properties of the annealed parent metal. This is no doubt a result of the highly acicular martensitic structure of the weld metal (Fig.3), with a large prior-beta grain size contributing an additional factor for the decrease in tensile elongation. This goes hand in hand with an increase in hardness in the fusion zone revealed in Fig. 7. Post-weld annealing at 650°C results in a reduction of strength and moderate increase in ductility from the as welded condition. Further decreases in tensile and yield

Table 5 : Fusion zone tensile properties of OT4-1 electron beam welds

No.	Condition	Yield strength (0.2% proof stress) (MPa)	UTS (MPa)	Elongation %
1.	As-welded	633	835	8.5
2.	PWHT 650°C	546	682	10.8
3.	PWHT 850°C	499	638	11.6
4.	PWHT 850°C/1h/WQ+540°C/16h/FC	513	653	11.2
	(PWHT+Post-weld heat treatment)			

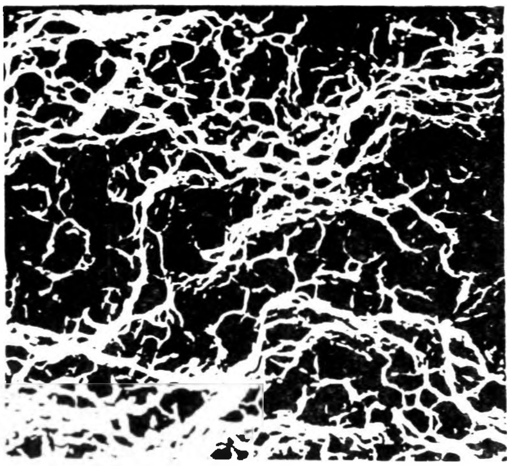

Fig. 8 : Scanning electron fractograph of tensile sample of the base metal (320X)

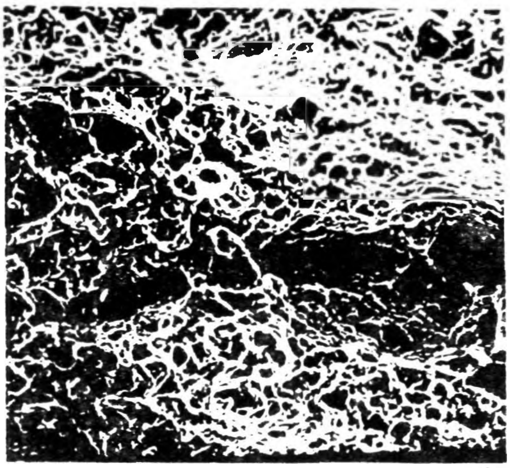

Fig. 9 : Scanning electron fractograph of fusion zone tensile sample in the as-welded condition (320X)

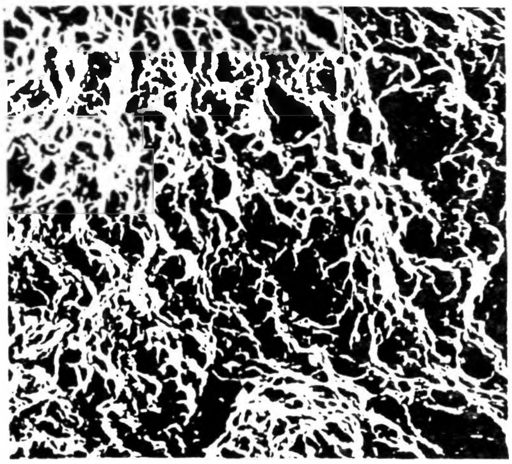

Fig. 10 : Scanning electron fractograph of fusion zone tensile sample after 650°C post-weld heat treatment (320X)

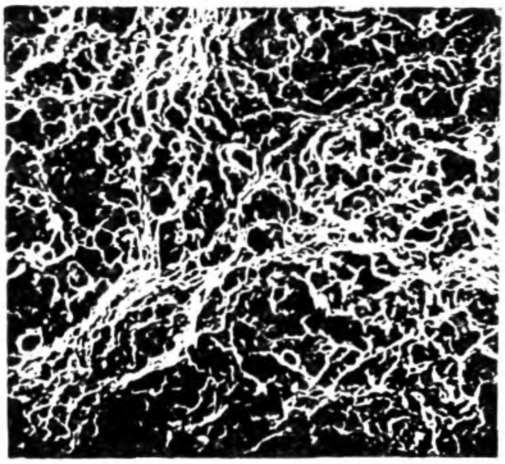

Fig. 11 : Scanning electron fractograph of fusion zone tensile sample after 850°C post-weld heat treatment (320X)

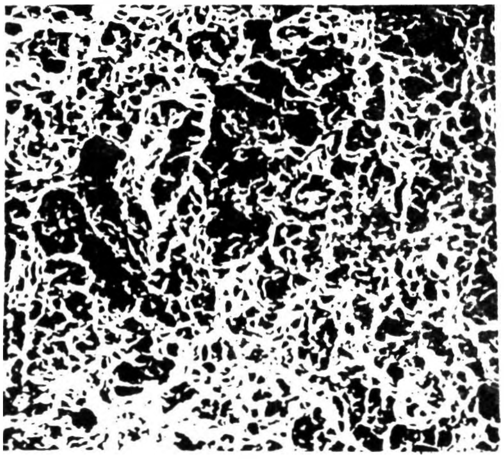

Fig. 12 : Scanning electron fractograph of fusion zone tensile sample after 850+540°C post-weld heat treatment (320X)

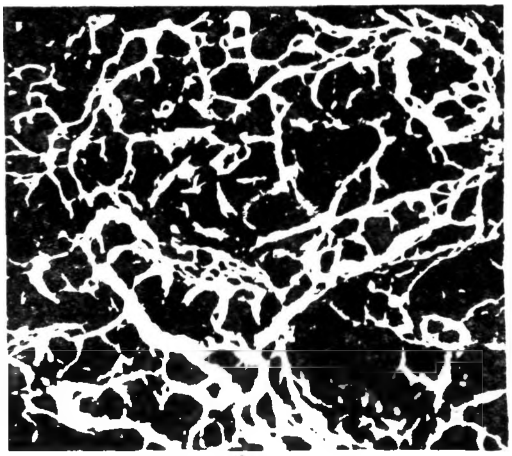

Fig. 13 : Scanning electron fractograph of fusion zone tensile sample in the as-welded condition (1280X)

strengths and improvements in ductility are obtained after the single post-weld heat treatment at 850°C for 3 hours and the duplex treatment of 850°C/1h/WQ+540°C/16h/FC. The tensile properties are very nearly the same after these two thermal treatments. The hardness distributions in Fig. 7 also reveal the softening effect of all the three post-weld heat treatments. It must be noted, however, that even after a three-hour anneal at 850°C, which is just 70°C below the beta transus, the tensile elongation of the weld metal annealed at the same temperature (15.5%, Table 2). Similarly, the fusion zone ductility after post-weld heat treatment at 650°C 3h (10.8%, Table 5) is lower than for the base material treated under an identical condition (14%, Table 2).

These results can be explained on the basis of the microstructures shown in Figs. 1-6. The high strength of the weld metal is obviously to be attributed to the sharply acicular martensitic structure, while its relatively low ductility is due to the dominant presence of martensite as well as the comparatively large prior-beta grain size. Post-weld heat treatment leads to a tempering of the martensite. In alpha-beta alloys this is believed to occur by the formation of the beta phase predominantly at martensite plate boundaries; this impoverishes the martensite in the beta-stabilising elements so that it degenerates to equilibrium alpha [12]. Another reaction that occurs during post-weld thermal treatment is the aging of any retained beta phase; this leads to alpha phase precipitation in the beta phase as the alloy seeks to attain thermodynamic equilibrium.

Some of these reactions can start occuring even at temperatures as low as 200°C, even though they are admittedly very sluggish at the low temperature and require many thousands of hours [8]. The kinetics are much more rapid at elevated temperatures and, as mentioned earlier, post-weld heat treatments lasting not more than a few hours have been reported over a wide range of temperatures. The tempering of the martensite may be expected to result in soferning, a reduction of strength and an increase in ductility. On the other hand, the aging of the beta phase is likely to provide a strengthening effect consequent to precipitation. Prolonged heat treatment at elevated temperatures results in overaging and a general coarsening of the microstructure; this will doubtless contribute to a strength reduction while the ductility may be expected to increase.

The tensile properties listed in table 5 show that all the three thermal treatments have led to a substantial reduction in strength with an attendant increase in ductility. This indicates that, at the temperatures of heat treatment in the range 650-850°C, the Significant reaction is the tempering of martensite. On the other hand, the aging reactions involving the precipitation of beta phase from the martensite and alpha phase from the retained beta phase have not resulted in any hardening. This is probably because the amount of retained beta phase in the as-welded condition is very small. Another possible reason is that overaging occurs at the temperatures used for heat treatment namely 650-850°C. The micrographs in Fig. 5 (850°C/3h/FC) and Fig.6 (850°C/1h/WQ + 540°C/16h/FC) reveal significant coarsening effects, while the microstructure in Fig. 4 (650°C/3h/FC) continues to remain acicular with a high aspect ratio. However the 650°C heat treatment has resulted in softening (Fig. 7) as also a reduction in strength (Table 5) in relation to the as-welded condition. These observations suggest that overaging phenomena may just be commencing at temperatures at and just below 650°C.

A comparison of he single heat treatment (850°C/3H/FC) and the duplex heat treatment (850°C/1h/WQ+540°C/16h/FC) fails to reveal any major differnece either in microstructure or in mechanical properties between the two. The microstructures reveal that the single but longer treatment at 850°C (Fig.5) shows a greater degree of coarsening both as regards the grain boundary alpha phase and the intragranular alpha phase plates; while the duplex treatment leads to somewhat improved intragranular definition. The latter is perhaps the consequence of the prolonged aging at 540°C (for 16h) involved in the duplex treatment. Notwithstanding these minor differences, the mechanical properties after these two heat treatment schedules are nearly the same.

Fractography

The results of scanning electron microscopic examination are seen in Figs. 8-14. **Fig. 8**, taken from the tensile fracture face of the base metal annealed at 850°C, shows a typical transgranular ductile fracture consisting of deep, essentially equiaxed dimples. The fracture of the as-welded specimen **Fig. 9** and the post-weld heat treated samples **(Figs. 10-12)** are also transgranular and ductile, except that the as-welded speciment shows relatively flatter features with shallow dimples. Very little difference is observed among the three post-weld heat treated fracture faces. The micrographs taken at higher magnification also show that the tear ridges enclose smaller, more shallow dimples in the as welded condition **(Fig.13)** *compaed to the duplex post-weld* heat treated condition **(Fig.14)**. In spite of these minor differences, however, the scanning electron fractograph show that all the fractures are transgranular and have occured in a ductile manner by micro-void nucleation and coalescence.

It is significant to record that, even though post-weld annealing has improved the fusion zone ductility over the aswelded value, the tensile elongations are still much lower than for the unwelded base material. This is obviously attributable to the morphology of the alpha phase in the welded and unwelded conditions. The fusion zone, solidifying and cooling through the beta phase field, is characterised by an acicular structure (Fig. 3), which continues to remain lamellar (though with a reduced aspect ratio (Fig. 4-6) during post-weld annealing upto the highest temperature of 850°C. The alpha-beta processed base material, on the other hand, exihibits a high proportion of primary alpha phase with an *aquiaxed* morphology. As the void nucleation leading to tensile fracture is known to occur preferentially at alpha-beta interfaces [13], lamellar microstructures, with a much larger interface area than equiaxed structures, result in a lower tensile elongation.

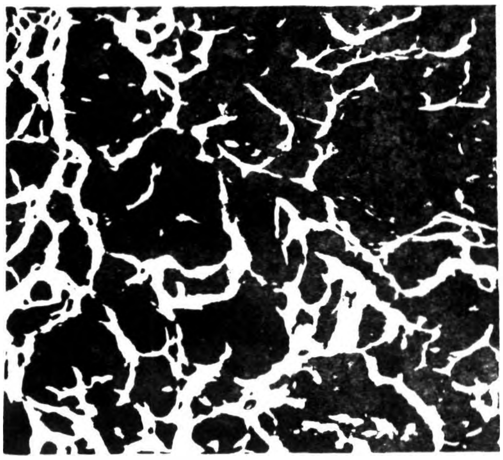

Fig. 14 : Scanning election fractograph of fusion zone tensile sample after 850+540°C post-weld heat treatment (1280X)

CONCLUSIONS

1. Electron beam welding of Ti-Al-Mn OT4-1 alloy yields weld metal of high strength but somewhat poor ductility in relation to the base metal on account of a needle-like martensitic strucutre and large prior-beta grain size.
2. Post-weld heat treatment in the range of 650-850°C improves ductility, though with some sacrifice in strength.
3. Even after a high-temperature post-weld heat treatment it is difficult to attain a fusion zone ductility matching that of the base material.
4. The post-weld heat treatment at 650°C leads to tensile properties not inferior to those after treatment at higher temperatures and hence will be adequate for practical application.

ACKNOWLEDGEMENT

The authors wish to thank the Department of Science and technology which sponsored the above work.

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