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Omega phase formation in RMI (38-6-44) beta titanium alloy

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The stabilized beta titanium alloy RMI (38-6-44) was designed to limit the formation of omega-phase during quenching and subsequent ageing. It has been shown in the present paper, however, that both athermal and thermal omega do form and that the growth kinetics of the thermal omega follows a $D = Ct^{0.45}$ relationship.

As expected from the alloy additions, the maximum volume fraction of omega phase observed is considerably less than in other beta titanium alloys and the severe embrittlement due to omega phase formation is not a serious problem in this alloy. This is an important consideration for engineering applications in the service temperature range below 900° F (480°C).

1. Introduction

The development of the stabilized beta titanium alloys has provided a group of light weight high strength alloys that can easily be worked at room temperature. In these alloys combinations of mechanical working and thermal treatments have been shown to give rise to a broad range of properties. However, the formation of a metastable transition phase, omega, during low temperature (<800°F, 420°C) ageing of beta phase alloys has been shown to produce severe embrittlement of the alloy. The reason for the embrittlement is still not clearly understood, although it has been reported [1] that, in the Beta III titanium alloy, the omega-phase tends to form preferentially at the beta-phase grain boundaries and forms a heavy grain-boundary envelope which apparently prevents the propagation of slip from one grain to the next and intergranular fracture occurs.

The RMI (38-6-44) beta titanium alloy was designed specifically to attempt to overcome some of the problems encountered in the other beta alloys such as, for example, excessive grain growth during solution heat-treatment and omega-phase formation during quenching and subsequent ageing.

The metallurgy of the RMI alloy is similar in character to the other beta titanium alloys, such that, solution treatment at 1500° F (810° C), which is just above the beta transus temperature of 1460° F (790° C), followed by quenching retains the body centred cubic beta phase at room temperature. The most significant difference between this alloy and other beta titanium alloys is that in the solution heat-treated condition the microstructure consists of small equiaxed beta grains with a finely dispersed secondary intermetallic phase ($TiCr_2$) in the beta matrix. The primary function of the dispersed phase, which forms during melting, is that of inhibiting grain growth during the solution treatment. Other alloying additions are made to suppress the formation of the omega phase.

The purpose of this present investigation was to determine if omega phase forms in the RMI beta alloy either on quenching from the solution treating temperature or during low temperature ageing.

2. Material and processing

The material for this investigation, supplied by Reactive Metals Incorporated, had a chemical analysis in wt % of Ti-3.4Al-7.8V-5.8Cr-4.1Mo-3.7Zr-0.090$_2$-0.16Fe-58 ppm H_2.

The material was received in the form of 0.25 in. diameter rod which was swagged to $\frac{1}{8}$ in. diameter at room temperature without intermediate annealing. After swagging, the material was solution heat-treated at 1500° F (420 °C) for

30 min in evacuated glass capsules which were then broken into water to produce a rapid quench. The specimens were chemically polished and aged in a nitrogen gas atmosphere at temperatures of 500, 600 and 700° F (260, 310 and 370 °C) for periods of 1, 5, 10, 24 and 48 h after which they were furnace-cooled to room temperature.

Specimens for electron microscopy were prepared by sectioning 0.010 in. thickness in the cross-section of the rods using a diamond wheel. These samples were then jet polished to thin foils using a methyl alcohol, butyl alcohol and perchloric acid solution at a temperature of −40° C. The foils were observed in a Hitachi HU125 electron microscope operated at 125 kV.

Transmission X-ray patterns were recorded on film from stationary thin specimen using filtered radiation from a silver target tube. X-ray patterns were obtained from solution treated and aged specimens.

3. Results and discussion

The selected-area diffraction pattern from the RMI titanium alloy after solution treatment at 1500° F (810 °C) and water quenching is shown in Fig. 1. There is pronounced intensity streaking in the $\langle 1\,1\,0\rangle_\beta$. This observation is similar to that reported by Min Chung Jon *et al* [1] for Beta III titanium and also by Sass [2] in Zr-25% titanium. Sass was able to show that the intensity streaks were due to discs of intensity normal to the $\langle 1\,1\,1\rangle_\beta$. Furthermore, through direct observation, the discs of intensity were shown to be associated with omega precipitates of about 50 Å in size which were aligned in rows in periodic arrays in the $\langle 1\,1\,1\rangle_\beta$ directions.

In the present studies, tilting the foils approximately ten degrees in the electron microscope did not eliminate the intensity streaks as would be expected if they were actually discs of intensity, nor was there any visible strain contrast associated with the omega particle contrast, as was also found to be the case in the Beta III work.

Diffuse diffracted intensity positioned inside the (1 1 0) beta phase Debye ring and streaking due to the existence of the omega phase was also present on the X-ray transmission diffraction patterns. The diffuse character of the intensity made positive identification of the diffracting planes impossible. However, an average spacing of 3.71 Å was computed from the diffuse ring. The $(10\bar{1}0)$ and (0 0 0 1) omega phase planes have

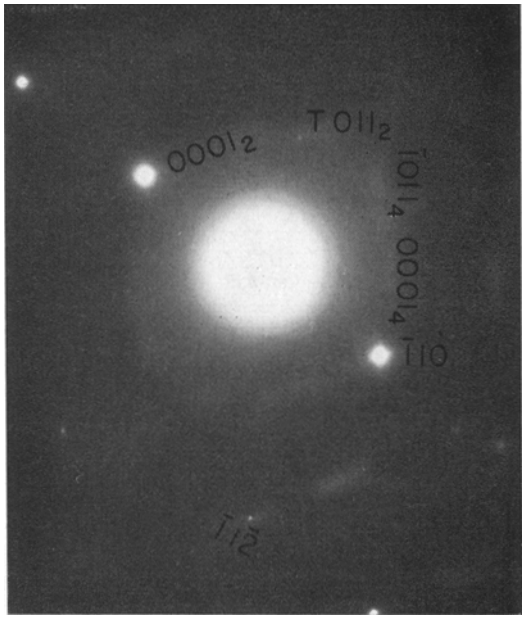


Figure 1 Selected-area electron transmission diffraction pattern from RMI (38-6-44) alloy in the solution treated condition, showing streaking in $\langle 1\,1\,0\rangle_\beta$ directions.

spacings of 3.975 and 2.815 Å [3]. The very diffuse nature of the diffracted intensity indicates that the omega-phase formed during the quench has a particle size less than 100 Å. From electron and X-ray diffraction effects it was assumed that the intensity streaks were caused by the presence of fine omega-phase particles aligned in the $\langle 1\,1\,1\rangle_\beta$.

It can be seen from the electron micrograph shown in Fig. 2 that upon ageing the quenched specimens for 5 h at 600° F (310 °C) the omega precipitates are easily resolved and have an image diameter of about 250 Å. The omega phase structure observed after ageing 48 h at 600° F is shown in Fig. 3. The particles now have an image diameter of about 700 Å. The precipitates, which are coherent with the matrix, were determined from selected area electron diffraction pattern analysis to have an orientation relationship as follows:

$$(0001)_\omega \,//\, (111)_\beta\,,\quad [2\bar{1}\bar{1}0]_\omega \,//\, [1\bar{1}0]_\beta\,,$$

which is the same as that reported by Silcock [3] and Bargaryatski [4]. The selected-area diffraction pattern, Fig. 3b, also shows intensity streaking in the $\langle 1\,1\,1\rangle_\beta$, and additional discrete intensity spots for omega.

Since there are four $\langle 1\,1\,1\rangle_\beta$, there should be

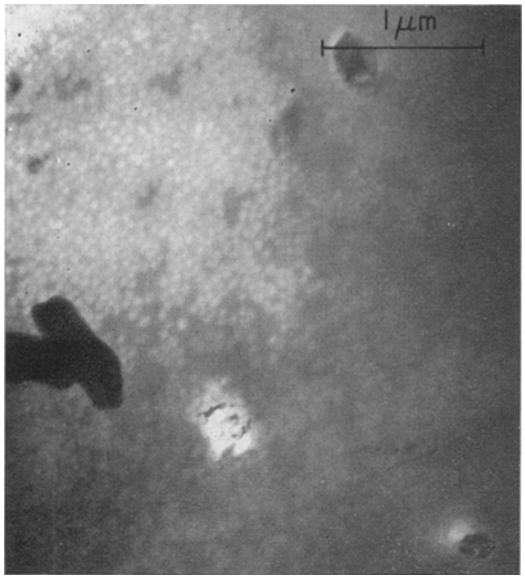


Figure 2 Resolved spherical shaped omega phase particles in RMI (38-6-44) alloy solution treated and aged 5 h at 600°F (310 °C).

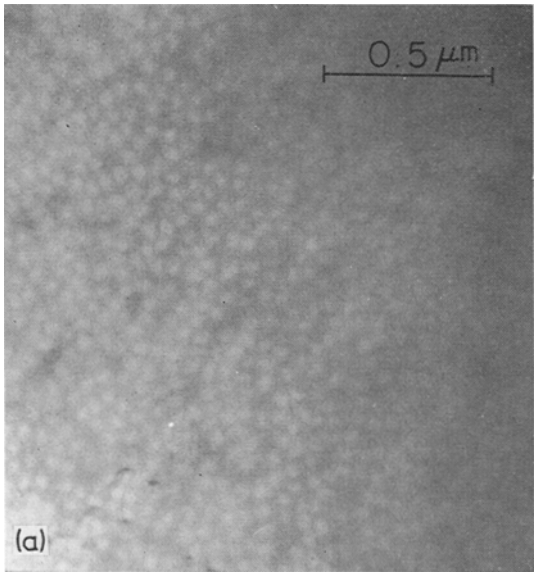
four orientations for the omega precipitates. This was clearly demonstrated in the Beta III work [1] where the particles were ellipsoidal and exhibited slightly different contrast in each of the different orientations. In the present studies, however, the particles appear to be spherical in shape and there was no contrast change to indicate the specific variant of any particle. The particle sizes observed after various ageing times are given in Table I. The maximum volume fraction of

TABLE I Particle size of omega phase, in RMI (38-6-44) alloy aged at 600°F (310°C)

Ageing time (h)	Particle size (Å)
5	250
24	500
48	700

omega phase observed in the RMI alloy after 48 h ageing at 600°F (310°C) was about 13.5%, this is considerably less than the maximum volume fraction observed in the Beta III titanium alloy by Min Chung Jon *et al* [1]. They reported more than 90% volume fraction of omega after ageing for 49 h at 725°F (380°C). However, the difference in the maximum volume fraction of omega formed in the two alloys could be accounted for by the fact that the electron to atom ratio of the alloying constituents in the RMI alloy is 4.19, as compared to 4.08 for the Beta III alloy, which is a little on the high side of the range of values of between 4.07 and 4.14 predicted by Luke *et al* [5] to be required for omega phase formation. It may be concluded, therefore, that the alloying constituents of the RMI alloy in fact tend to suppress but do not eliminate completely the formation of the omega phase on the quench or during low temperature ageing.

The particle size data presented in Table I were



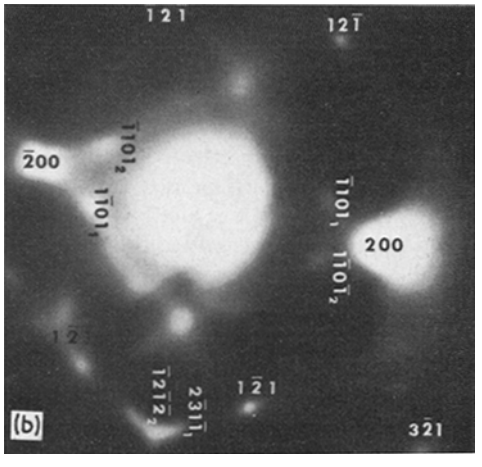


Figure 3 Omega phase particles in RMI (38-6-44) alloy solution treated and aged 48 h at 600°F (310°C). (a) bright-field image, (b) selected-area diffraction pattern from area in (a).

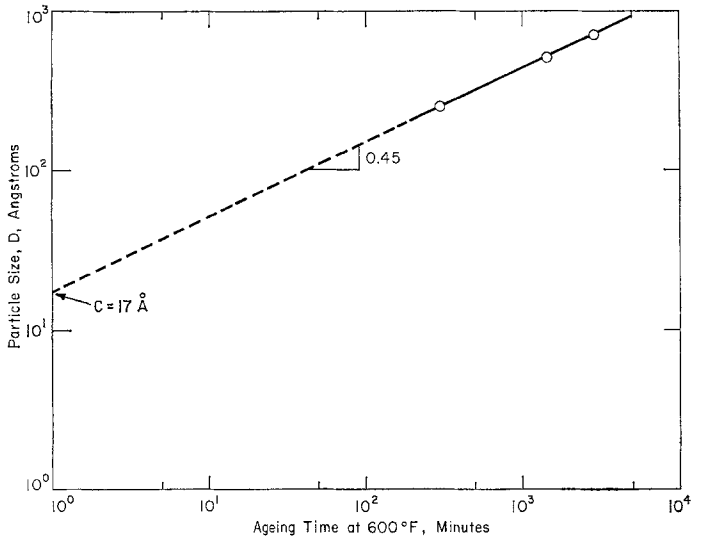


Figure 4 Omega phase particle size in RMI (38-6-44) as a function of ageing time at 600° F (310 °C).

fitted to the empirical growth law [6], $D = Ct^n$, where D is the particle diameter and t is the ageing time. These data are shown in Fig. 4, from which the value for the exponent n was found to be 0.45. Extrapolation of the line back to zero (ageing time of 1 min) gave an estimate of the initial particle size of about 20 Å, a size which would not be easily resolved in the electron microscope and would give the very diffuse X-ray scattering. The growth kinetics of the omega phase in the RMI alloy appears to be faster than those measured in the Beta III alloy, and the reported initial size at ageing time at 1 min of about 300 Å is considerably larger than the present case.

After solution treating and ageing 48 h at 600° F (310 °C), X-ray reflections from the omega phase appear as sharper spots and the general streaking of the reflections from the beta phase is eliminated. These two effects indicate the increase in the omega particle size and an increase in the length of perfect structure periodicity in the beta phase. However, after this ageing treatment the omega reflections were not sharp enough to determine the index of the reflecting planes.

In those specimens aged for 1 h at 700° F (370 °C) no omega phase could be observed by electron microscopy, although intensity streaking between diffraction spots could still be observed. At long ageing times at this temperature the precipitation and growth of alpha phase become predominant.

Acknowledgements

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