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Phase transformations in Ti–V alloys

Part 2 *α and ω formation*

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The precipitation of the α-phase during isothermal transformation in the (α + β) phase field has been studied in the Ti–10V and the Ti–20V alloys. It has been found that the α-phase emerges as Widmanstatten plates and the subsequent quench after the isothermal transformation induces the β → ω displacive reaction in the matrix β-phase. The formation of the ω-phase during ageing is also reported in this paper and the influence of prior β-quenching on α precipitation has been studied in detail in the Ti–20V alloy. Some observations on the type 2 or non-Burgers' α crystals are also discussed in this paper.

1. Introduction

Metastable β titanium alloys are currently receiving a great deal of attention [1–8] mainly due to their amenability to strengthening heat treatments. The metastable β titanium alloys can be strengthened to very high levels by precipitation of the hcp α-phase in the bcc β matrix. Formation of transition phases such as ω-phase in solute lean alloys or the $\beta + \beta'$ structure in solute-rich alloys can also be considered for potentially strong microstructures. The morphology of the α-phase can be drastically changed by selection of one of the following heat-treatment sequences: (i) β-quenching followed by ageing; and (ii) isothermal holding in the $(\alpha + \beta)$ phase field subsequent to β-solutionising. It has been reported that the alloys subjected to the latter treatments exhibit strength and ductility combinations and impact properties generally superior to those obtained by quenching and ageing [9]. The α-phase has been found to precipitate in the form of Widmanstatten plates subsequent to this treatment and this structure is associated with high strength and good fracture toughness [10]. Widmanstatten microstructure has also been found to possess superior resistance to fatigue crack growth in Ti–6Al–4V [11]. However, the structure is associated with poor ductility. Refinement of the microstructure to produce a fine distribution of the α-phase in the β matrix by thermomechanical treatments or a combination of these two has been shown to enhance the tensile ductility [12–15].

The α-phase can precipitate in the β matrix, it can precipitate heterogeneously on the ω particles present initially in a β-quenched material or it can evolve as the stable, equilibrium precipitate in the β matrix from a $(\beta + \omega)$ microstructure subjected to prolonged holding at temperatures where the ω-phase can exist in metastable equilibrium. The morphology and distribution of the α-phase would thus depend strongly on the path followed in producing the microstructure.

Conventionally, the presence of the metastable, hexagonal non-close packed ω-phase is associated with a dramatic loss of ductility although the alloys are strengthened to a considerable extent [16–20]. However, recently, many attempts have been made by several workers in order to circumvent the embrittlement problem associated with the ω formation by optimizing the volume fraction of the ω-phase or by a short-time ageing of the ω-embrittled material at a higher temperature so as to cause heterogeneous precipitation of the α-phase on these ω particles [12, 14, 21–24].

From these studies it appears that the selection of the ageing cycle to which a metastable β titanium alloy is subjected should primarily be based on the knowledge of the transformation characteristics of

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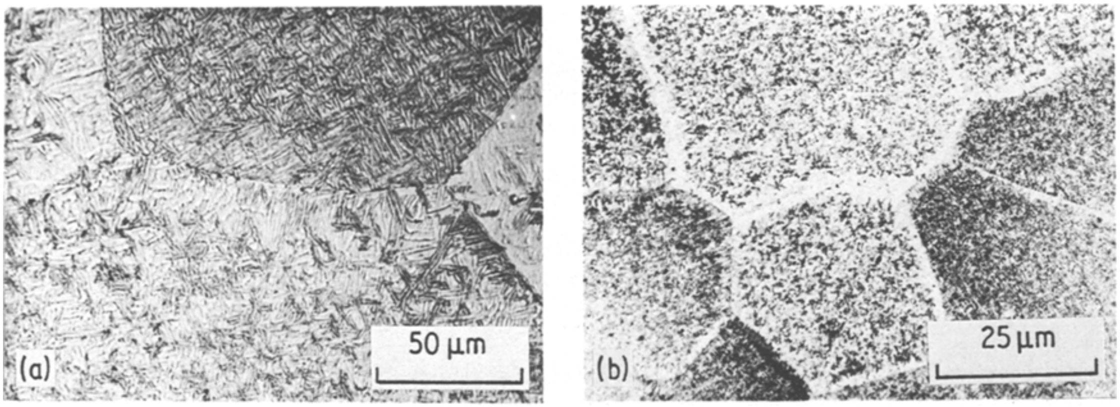


Figure 1 (a) Ti–10V, β-solutionized and isothermally reacted at 973 K for 1 h. Optical micrograph showing grain-boundary allotriomorphs and Widmanstatten plates. (b) Ti20V, β-solutionized and isothermally reacted at 823 K for 1000 min. Optical micrograph showing fine α precipitates in a β matrix.

the alloy. In this study, various microstructures have been produced by heat-treating Ti–10V and Ti–20V alloys in the $(\alpha + \beta)$ phase field and an attempt has been made to characterize these microstructures.

2. Experimental procedure

Most of the experimental details have been described in Part 1 [25]. All isothermal transformation treatments were carried out by employing a tandem arrangement of furnaces whereby the β-solutionized specimens could be quickly transferred to the ageing furnace. The Ti–10V samples were β-solutionized by holding them at 1273 K for 15 min, isothermally reacted at 973, 948, 923, 898, 873 and 823 K and held for 1 h, and subsequently water-quenched. The Ti–20V samples were held at 1223 K for 15 min and rapidly transferred into the furnace maintained at 823 K for different periods of time. In another set of experiments, the Ti–20V samples were β-quenched and aged at 823 K for different periods of time.

3. Results and discussion

The isothermal decomposition in the $(\alpha + \beta)$ phase field subsequent to β solutionization led to the emergence of the primary α-phase. As the reaction temperature in the $(\alpha + \beta)$ phase field was lowered, the volume fraction of the α-phase increased. The matrix presumably became correspondingly enriched in the solute element. The morphological and the crystallographic aspects of the α-phase in these samples are discussed in the following sections.

3.1. Morphology

3.1.1. Morphology of the α-phase in isothermally treated samples

The α-phase was found to nucleate at the grain boundaries of the matrix β-phase and grow preferentially and more or less smoothly along the grain boundaries. The interior of the grain was found to have a distribution of the Widmanstatten α plates. As can be seen from Fig. 1, the α plates in Ti–20V samples were considerably smaller in size when compared to those formed in the Ti–10V samples. With the lowering of the reaction temperature, finer α plates were observed in the Ti–10V samples too. The morphological observations are in general agreement with similar studies on various titanium-base alloys [26–28].

In both of these alloys, a major volume fraction of the α-phase was present as Widmanstatten plates. The α plates formed in these specimens were invariably associated with layers of fine crystals of the α-phase at the α/β interfaces (Fig. 2a and b). The formation of fine contiguous particles of the α-phase around the monolithic α plates has been observed in various titanium- and zirconium-base alloys [27, 29–33]. As illustrated in Fig. 2, the α plates appeared to have a complex interface structure. SAD patterns obtained from these areas revealed that arcing was associated with the bcc reflections (Fig. 3). Dark-field microscopy revealed that the interface phase is not continuous, but comprised small, individual crystallites. This is best seen in Fig. 4 which shows fine α crystals enveloping individual α plates. This observation strongly suggests that during the growth of the plates, fresh nucleation of the precipitate crystals

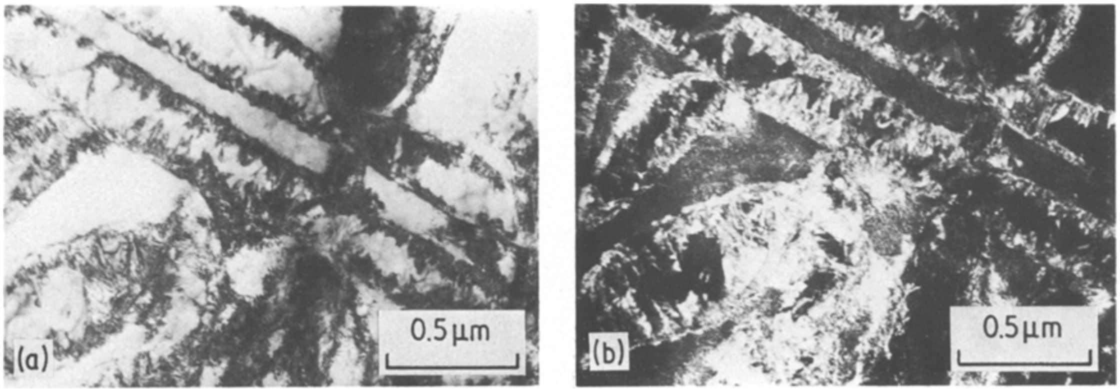


Figure 2 Ti–10V, β-solutionized and isothermally reacted at 973 K for 1 h. (a) Bright- and (b) dark-field micrograph, respectively, showing the typical appearance of the type 1 and the type 2 α-phases. The dark-field micrograph was obtained with an arced reflection.

occurred at the faces of the previously formed plates of the same phase. Such a process has been termed *sympathetic nucleation* [26]. This will be discussed in detail in a subsequent section.

3.1.2. Morphology of the α plates in quenched and aged samples

As mentioned in Part 1 [25], water quenching the Ti–20V alloy retained the high-temperature β-phase in a metastable manner. However, this quenching did not suppress the $\beta \rightarrow \omega$ reaction and the quenched alloy contained a distribution of fine unresolvable ω particles in the β matrix. During ageing of these β-quenched specimens, the ω particles presumably acted as nucleation sites and hence the precipitation reaction was found to be greatly accelerated. Accordingly, in samples aged for very short periods, a uniform distribution of fine coherent α-phase was noticed. SAD patterns obtained from these specimens showed diffraction effects associated with the ω-phase. Ageing these specimens for long periods was found to have resulted in the growth of the α-phase into large plate or needle-shaped precipitates (Fig. 5a and b). It should be pointed out that these α plates (or needles) are much smaller in size when compared with the α plates formed during isothermal ageing. However, the morphological features remained identical and the interfacial α-phase was observed in all these specimens.

3.2. Crystallography and orientation relationships

SAD patterns from the samples containing a distribution of Widmanstatten α plates showed that the α-phase was related to the matrix β-phase through the Burgers' bcc/hcp orientation relationship. A typical reciprocal lattice section, $[1\,1\,1]_\beta$,

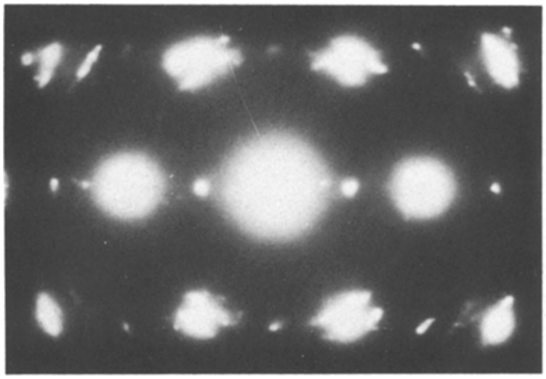


Figure 3 SAD pattern obtained from Ti–10V, β-solutionized and isothermally reacted at 823 K for 1 h. Zone axis: $[1\,1\,1]_\beta$.

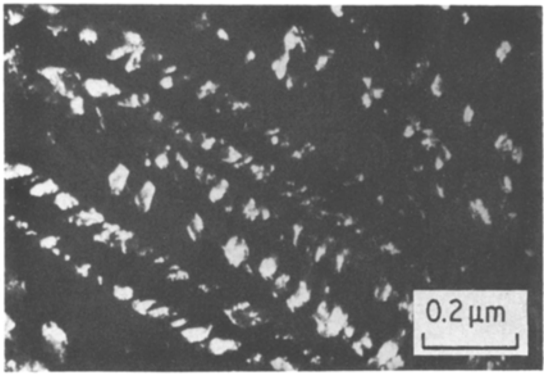


Figure 4 Ti–10V, β-solutionized and isothermally reacted at 823 K for 1 h. Dark-field micrograph obtained from an arced reflection showing the enveloping α crystals.

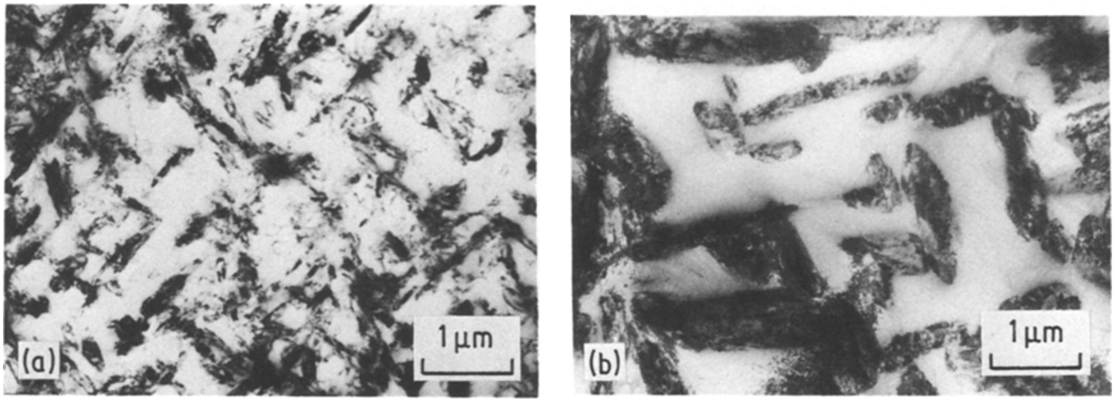
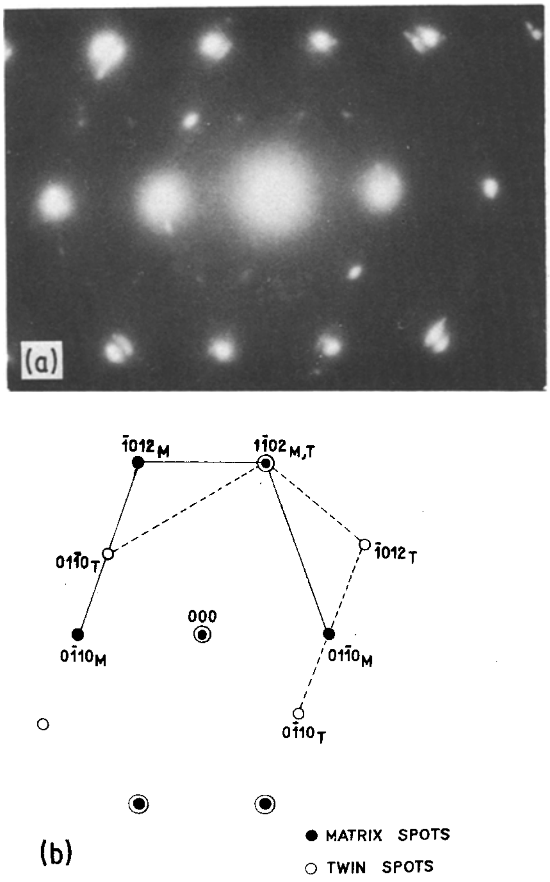


Figure 5 Ti–20V, β-quenched and aged at 823 K for (a) 100 min and (b) 1000 min. These bright-field micrographs illustrate the typical appearance of the α-phase in quenched and aged samples.

obtained from an alloy studies here is shown in Fig. 3. It can be seen that in addition to the type 1 α-phase (or Burgers' α, characterized by the $(10\bar{1}0)_\alpha$ reflections between the $(110)_\beta$ reflections in the $[111]_\beta$ section, Fig. 3), the type 2 α-phase characterized by the arced reflections, is also present in these alloys. It was found that the type 2 α-phase and the matrix β-phase were not related through Burgers' orientation relationship and it has been referred to in the literature as the non-Burgers' α-phase [29]. Interestingly, it was observed in this study that the type 2 α crystals are related to the type 1 α-phase through $\{10\bar{1}2\}_\alpha$ twinning. This is illustrated in Fig. 6. Rhodes and Williams [29] have deduced from stereographic analyses that such a crystallographic relation exists between the two types of α crystals.

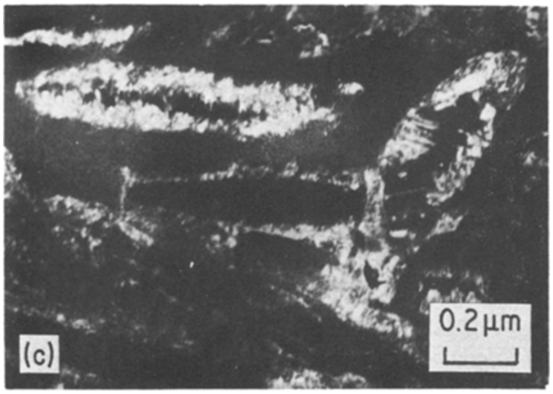


3.3. Mechanism of formation of type 2 α crystals

Two possible mechanisms of formation of the type 2 α-phase have been discussed by Rhodes and Williams [29]: (i) mechanical twinning of type 1 α; and (ii) nucleation and growth of the type 2 α to replace type 1 α.

Twinning is a well-known mode of defor-

Figure 6 Ti–10V, β-solutionized and isothermally reacted at 973 K for 1 h. (a) SAD pattern and (b) key showing $\{10\bar{1}2\}_\alpha$ twin relation. (c) Dark-field micrograph obtained with a $(01\bar{1}0)_T$ reflection.



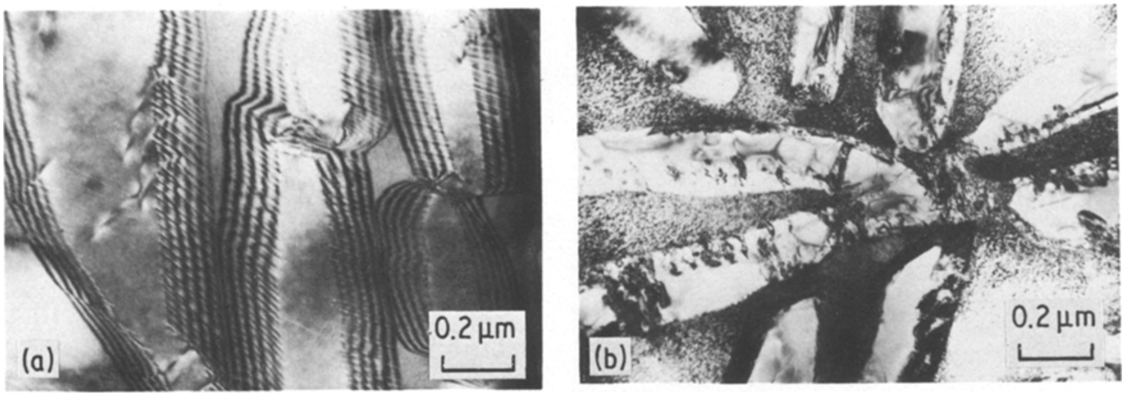


Figure 7 Ti–10V, β-solutionized and isothermally reacted at (a) 873 K for 1 h showing the typical α/β interface and (b) 948 K for 1 h showing predominantly type 1 α-plates distributed in a β–ω matrix.

mation in titanium and zirconium alloys and $\{10\bar{1}2\}\langle 10\bar{1}1\rangle$ twinning is known to have the smallest twinning shear [34, 35]. Mechanical twinning of the type 1 α precipitates can occur during cooling due to volumetric constraints and differences in thermal expansion. However, the formation of the type 2 α along the perephery of the type 1 α is not explained by this mechanism as pointed out by Rhodes and Williams [29].

Rhodes and Williams [29] have proposed another mechanism by which the compositional difference between the type 1 α- and type 2 α-phases provides the driving force for the nucleation of the type 2 α-phase. However, the compositional difference of the two types of α-phases has not been demonstrated. According to them, the type 2 α-phase grows and consumes the type 1 α-phase in the course of time. In the present study, however, prolonged ageing (1000 min) of the Ti–10V alloy at 973 K was not found to change the morphology or the relative proportion of the two types of α-phases.

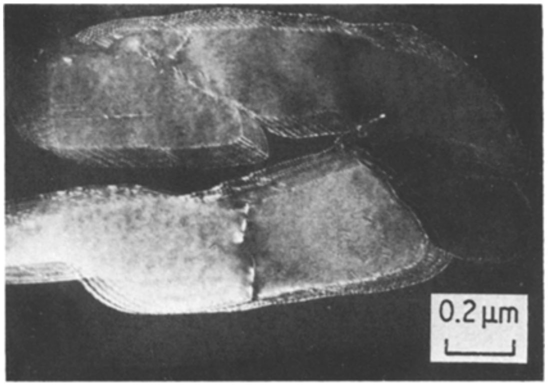


Figure 8 Ti–10V, β-solutionized and isothermally reacted at 873 K for 1 h. Dark-field micrograph illustrating that three α crystals are present in what appears to be a single crystal.

Another interesting observation made in the present work is that in contrast to the morphology of the α plates formed at 973 and 823 K, those precipitated at 873 K were found to be largely free of the type 2 α interface phase. The majority of the α plates formed here showed sharp interfaces and the normal interfacial dislocations and fringe contrast (Fig. 7a). A small fraction of the type 2 α-phase was found to be associated with the type 1 α plates formed at 948 and 898 K (Fig. 7b). The Burgers' orientation relationship was found to be valid for these plates. The presence of small-angle boundaries separating crystals of nearly identical orientation was frequently noticed (Fig. 8). It was possible that the Widmanstatten plates were not monolithic single crystals but consisted of low misorientation arrays of contiguously located crystals. Such a morphology of the α-phase was suggestive of the operation of the *sympathetic nucleation* mechanisms [26, 36]. Since the formation of an α crystal would involve the rejection of V atoms from the region where the precipitate forms to the neighbouring β regions, the occurrence of sympathetic nucleation would be difficult from the considerations of chemical free energy alone. However, sympathetic nucleation of α crystals on the surface of pre-existing α plates would be possible if the specific surface energy of the α/α interface were much smaller than that of the β/α interface. The α/α interface energy would be low if the boundaries separating the continguous α crystals were small-angle boundaries as was in the present case. As pointed out by Rhodes and Williams [29], the coherent $\{10\bar{1}2\}_\alpha$ twin bound-

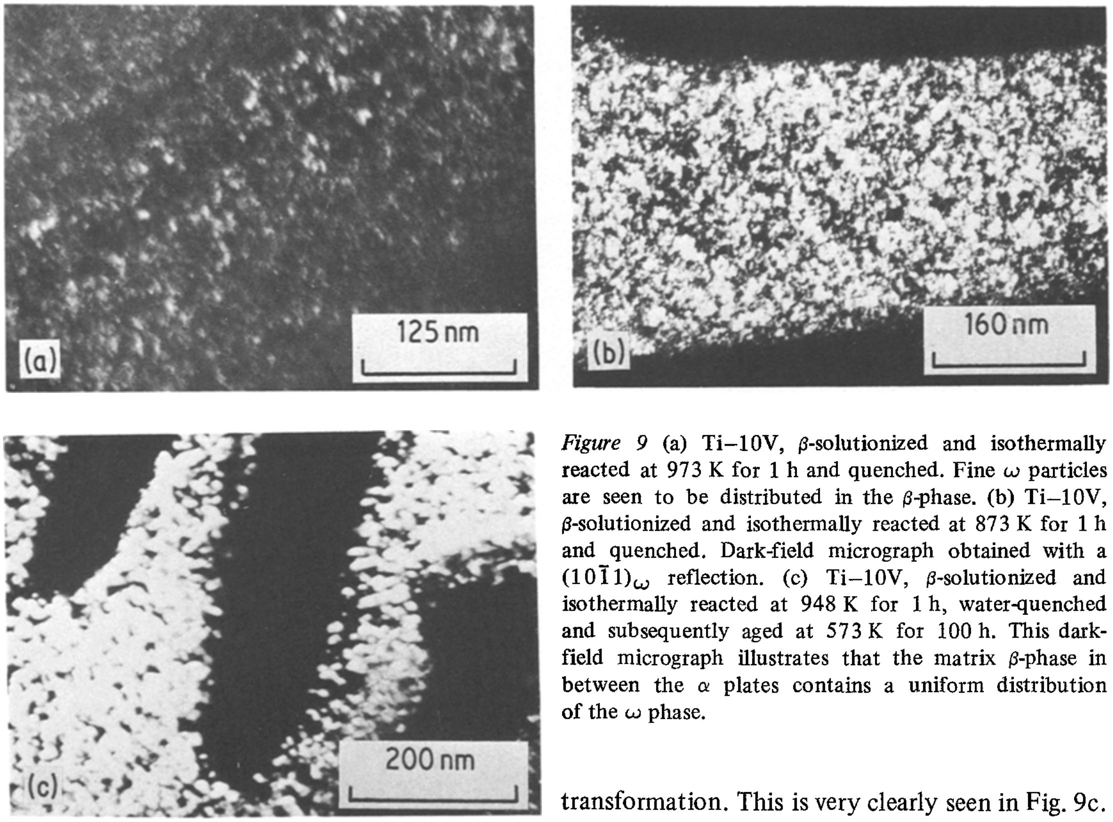


Figure 9 (a) Ti–10V, β-solutionized and isothermally reacted at 973 K for 1 h and quenched. Fine ω particles are seen to be distributed in the β-phase. (b) Ti–10V, β-solutionized and isothermally reacted at 873 K for 1 h and quenched. Dark-field micrograph obtained with a $(10\bar{1}1)_\omega$ reflection. (c) Ti–10V, β-solutionized and isothermally reacted at 948 K for 1 h, water-quenched and subsequently aged at 573 K for 100 h. This dark-field micrograph illustrates that the matrix β-phase in between the α plates contains a uniform distribution of the ω phase.

aries are associated with a low interfacial energy and so it appears reasonable to postulate that the type 2 α-phase formed as a result of sympathetic nucleation of α crystals at the broad faces of pre-existing α plates with the $\{10\bar{1}2\}$ interfaces separating them.

3.4. ω formation

During isothermal decomposition of these alloys in the $(\alpha + \beta)$ phase field, the β phase becomes progressively solute rich. The ω start temperature is known to show composition dependence in a manner similar to the martensitic start temperature [37, 38]. Isothermal experiments thus provide a very easy means of studying the effect of composition on the nature of the $\beta \rightarrow \omega$ transformation. The ω-phase formed in Ti–10V samples transformed at 973 K appeared as very fine paticles (Fig. 9a) while relatively larger ω particles were found to be present in the sample quenched from 873 K and there was a substantial increase in the number density of these particles (Fig. 9b). These ω particles were dispersed uniformly in the β-phase regions that separated the adjacent α plates which were formed during the isothermal transformation. This is very clearly seen in Fig. 9c. The Ti–10V and the Ti–20V samples subjected to $(\alpha + \beta)$ treatment by 823 K showed that the $\beta \rightarrow \omega$ transformation did not occur to any significant extent in these samples during quenching. This is best illustrated by the SAD patterns shown in Fig. 10. This is in accordance with the behaviour of the ω_s temperature which is lowered by higher solute content in the β-phase [37–39].

4. Conclusions

(1) The α-phase emerges as grain-boundary allotriomorphs and Widmanstatten side plates during isothermal decomposition in the $(\alpha + \beta)$ phase field. By quenching and ageing, a uniform distribution of plates of the α-phase could be obtained.

(2) Two types of α-phase, one obeying the Burgers' bcc–hcp orientation relationship and the other related to these α crystals through $\{10\bar{1}2\}$ twin relation, were observed in these alloys. It has been suggested that the type 2 α crystals nucleate sympathetically at the type 1 α plate interfaces.

(3) Depending upon the ageing temperature, the matrix β-phase was found to undergo a partial $\beta \rightarrow \omega$ displacive transformation. The ω particles present in the β-phase acted as heterogeneous nucleation sites for the α-phase and a uniform

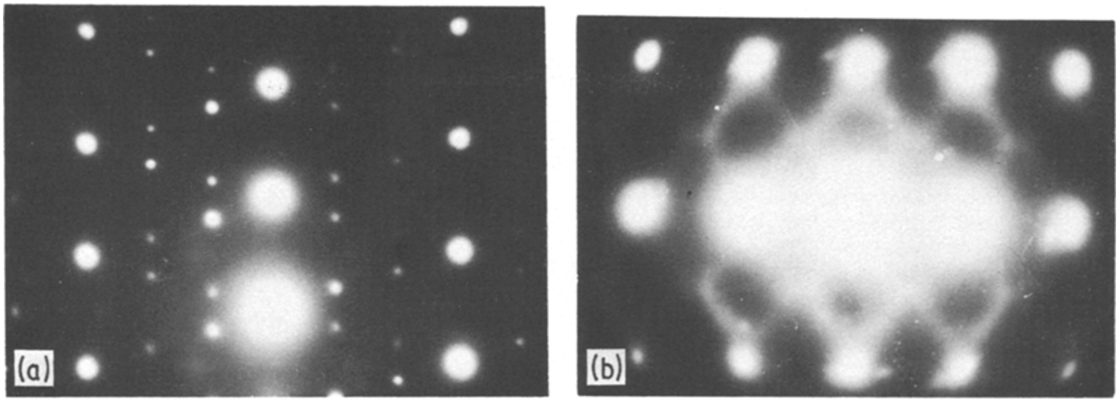


Figure 10 (a) Ti–10V, β-solutionized and isothermally reacted at 873 K for 1 h and water quenched. $[113]_\beta$ pattern showing four variants of the ω phase. (b) Ti–20V, β-quenched. $[110]_\beta$ pattern showing diffuse ω streaks.

distribution of the α-precipitates could be obtained by a quenching and ageing treatment.

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References

1. J. B. GUERNSEY, *SAE Trans.* **29** (1970) 2540.
2. D. N. WILLIAMS, R. WOOD and E. S. BARTLETT, *Met. Trans.* **3** (1972) 1529.
3. E. BOHANEK, in "Titanium Science and Technology", Vol. 3, edited by R. I. Jaffee and H. M. Burte (Plenum Press, New York, 1973) p. 1993.
4. R. CHAIT and C. STEAD, *Met. Trans.* **8B** (1977) 371.
5. C. C. CHEN and R. R. BOYER, *J. Metals.* **31** (1979) 33.
6. F. H. FROES, C. F. YOLTON, J. M. CAPANOS, M. G. H. WELLS and J. C. WILLIAMS, *Met. Trans.* **11A** (1980) 21.
7. T. W. DUERIG, G. T. TERLINDE and J. C. WILLIAMS, *ibid.* **11A** (1980) 1987.
8. R. CHAIT and T. S. DESISTO, in "Titanium Science and Technology", Vol. 2, edited by R. I. Jaffee and H. M. Burte (Plenum Press, New York, 1973) p. 1377.
9. D. J. DE LAZARO and W. ROSTOKER, *Trans. ASM* **46** (1954) 292.
10. D. B. HUNTER and S. V. ARNOLD, in "The Science, Technology and Applications of Titanium", edited by R. I. Jaffee and N. E. Promisel (Pergamon Press, Oxford, 1970) p. 959.
11. G. R. YORDER, L. A. COOLEY and T. W. CROOKER, *Eng. Fract. Mech.* **11** (1979) 805.
12. L. A. ROSALES, K. ONO, L. A. LEE and A. W. SOMMER, in "Titanium Science and Technology", Vol. 3, edited by R. I. Jaffee and H. M. Burte (Plenum Press, New York, 1973) p. 1813.
13. R. R. BOYER and P. T. FINDEN, *Met. Trans.* **3** (1972) 744.
14. A. W. BOWEN, *J. Mater. Sci.* **12** (1977) 1355.
15. G. M. PENNOCK, H. M. FLOWER and D. R. F. WEST, in "Titaniuum '80", Vol. 2, edited by H. Kimura and O. Izumi (AIME, Warrendale, Pennsylvania, 1980) p. 1343.
16. A. R. TROIANO and EL. H. HARMON, *Trans. ASM* **53** (1961) 43.
17. F. R. BROTZEN, EL. H. HARMON and A. R. TROIANO, *J. Metals.* **7** (1955) 413.
18. P. GANESAN, R. J. DE ANGELIS and G. A. SARGENT, *J. Less-Common Metals* **34** (1974) 209.
19. M. K. KOUL and J. F. BREEDIS, *Met. Trans.* **1** (1970) 1451.
20. C. E. ELLS and C. D. WILLIAMS, *Trans. AIME* **245** (1969) 1321.
21. M. K. KOUL and J. F. BREEDIS, *Acta Metall.* **18** (1970) 579.
22. M. J. BLACKBURN and J. C. WILLIAMS, *Trans. AIME* **242** (1968) 2461.
23. J. C. WILLIAMS, B. S. HICKMAN and H. L. MARCUS, *Met. Trans.* **2** (1971) 1913.
24. R. R. BOYER, R. TAGGART and D. H. POLONIS, *Metallogr.* **7** (1974) 241.
25. E. S. K. MENON and R. KRISHNAN, *J. Mater. Sci.* **18** (1983) 365.
26. H. I. AARONSON, in "Decomposition of austenite by diffusional processes", edited by V. F. Zackay and H. I. Aaronson (Interscience, New York, 1962) p. 387.
27. M. UNNIKRISHNAN, E. S. K. MENON and S. BANERJEE, *J. Mater. Sci.* **13** (1978) 1401.
28. H. IKAWA, S. SHIN, M. MIYAGI and M. MORIKAWA, in "Titanium Sciences and Technology", Vol. 3, edited by R. I. Jaffee and H. M. Burte (Plenum Press, New York, 1973) p. 1545.
29. C. G. RHODES and J. C. WILLIAMS, *Met. Trans.* **6A** (1975) 2103.
30. C. G. RHODES and N. E. PATON, *ibid.* **8A** (1977) 1749.
31. M. I. ZAKHAROWA, S. A. KIROV and A. G. KHUD-

ZHUA, *Phys. Stat. Sol. (a)* 49 (1978) 803.

32. D. O. NORTHWOOD, *J. Less-Common Metals* 61 (1978) 199.
33. M. UNNIKRISHNAN, M.Sc. Thesis, Bombay University (1976).
34. P. G. PARTRIDGE, *Met. Rev.* 118 (1967) 169.
35. R. E. REED-HILL, in "Deformation Twinning" Metals Society Conference, Vol. 25, edited by R. E. Reed-Hill, J. P. Hirth and H. C. Rogers (Gordon and Breach, New York, 1964) p. 295.
36. H. I. AARONSON and C. WELLS, *J. Metals.* 8 (1956) 1216.
37. S. L. SASS, *J. Less-Common Metals* 28 (1972) 157.
38. B. A. HATT, J. A. ROBERTS and G. I. WILLIAMS, *Nature* 180 (1975) 1046.
39. D. J. COMETTO, G. L. HOUZE, Jr and R. F. HEHEMAN, *Trans. AIME* 233 (1965) 30.

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