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The Effect of α Phase on the Deformation Mechanisms of β Titanium Alloys

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The effects of α and β phase interactions on the tensile and creep deformation behavior of β titanium alloys was studied using Ti-6.0wt.%Mn and Ti-8.1wt.%V as the model two-phase alloys, and Ti-13.0wt.%Mn and Ti-14.8wt.%V as the single-phase β alloys. The β phase of α-βTi-8.1V deforms by stress-induced hexagonal martensite (α′), while slip and twinning occurs in the single-phase β alloy with the same chemistry as the β phase. No stress-induced martensite was observed in the β or α-βTi-Mn alloys. This behavior is modeled in terms of a number of factors, including elastic interaction stresses between the α and β phases, coherency between the α phase and hexagonal martensite, and β phase stability.

Keywords elastic interaction stresses, stress-induced martensite, titanium, transmission electron microscopy, twinning

1. Introduction

Beta titanium alloys are technologically important. The deformation mechanisms of these alloys depend on a number of factors including grain size and β phase stability. It is important to understand the deformation mechanisms to design a desired microstructure to optimize properties. The deformation mechanisms of single-phase β alloys include fine slip, coarse slip, twinning, and the formation of stress-induced martensite or stress-induced ω phase (Ref 1, 2). The deformation mechanisms of single-phase α titanium alloys are fine slip, coarse slip, and twinning (Ref 1). Two-phase α + β alloys can deform by the mechanisms listed for the individual phases above, as well as by interphase-interface sliding (Ref 3).

In regard to the Widmanstätten α + β microstructures, it is well known that a Burgers orientation relationship exists between the α and β phases (i.e., slip systems in α and β are parallel) (Ref 4-9). Slip can be transmitted across the α-β interface from a slip system in one phase to another in the adjacent phase (Ref 10). However, there are no systematic studies that have related the deformation mechanisms of individual phases to those of the two-phase materials. This is not a simple extension of the behavior of single-phase materials because additional interactions occur between phases. These include elastic interaction between phases, localized stress from deformation products in adjacent phases, and α phase templating due to the Burgers orientation relationships mentioned above. In this regard, the present investigation addresses the effect of the α phase on the β phase deformation mechanisms in α-β Widmanstätten titanium alloys, which can differ from the deformation mechanisms of the β phase when present as a single-phase alloy.

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Table 1 Chemical compositions of the α, β, and α-β alloys used in this study

Alloy	Microstructure	α phase composition	β phase composition
Ti-0.4wt.%Mn	~100% α	~Ti-0.4wt.%Mn	N/A
Ti-6.0wt.%Mn	~46% α, ~54% β	~Ti-0.4wt.%Mn	~Ti-13.0wt.%Mn
Ti-13.0wt.%Mn	~100% β	N/A	~Ti-13.0wt.%Mn
Ti-1.5wt.%V	~100% α	~Ti-1.6wt.%V	N/A
Ti-8.1wt.%V	~51% α, ~49% β	~Ti-1.6wt.%V	~Ti-14.8wt.%V
Ti-14.8wt.%V	~100% β	N/A	~Ti-14.8wt.%V

Note: N/A, not applicable

2. Experimental Procedure

Single- and two-phase Ti-V and Ti-Mn alloys were selected such that the chemistry of the single-phase α and β alloys would match the component α and β phases of the two-phase alloys. Table 1 shows the chemistry and microstructure of these alloy systems.

The alloys shown in Table 1 were processed by the procedure given in Ref 11. When the alloys are processed in this way they will recrystallize in about 12 h at a temperature of 650 °C (Ref 12, 13). To obtain a Widmanstätten α + β microstructure, the alloys were annealed in quartz tubes sealed at 10^{-5} to 10^{-6} torr at 900 °C for 0.5 h, then they were furnace-cooled to 690 °C and annealed for 200 h, followed by water quenching. Tensile specimens were machined from the bars, with flats cut onto the gage section by electric discharge machining. The flat gage sections were mechanically polished, electropolished, etched, and fixed with gold fiducial lines 1 to 2 μm wide that were spaced 20 μm apart (Ref 14). This grid served as a reference to identify deformation mechanisms and to aid in the identification of interphase interface sliding.

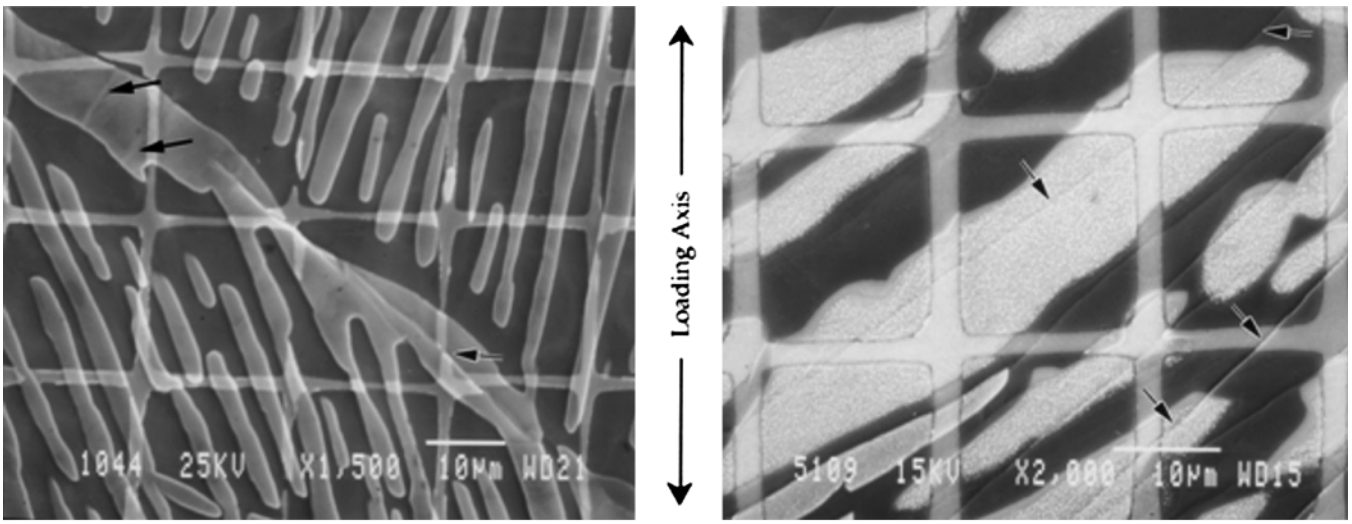

Fig. 1 SEM micrograph of (a) Ti-6.0wt.%Mn alloy and (b) Ti-8.1wt.%V alloy after tensile deformation to 3% of total strain. Arrows indicate coarse deformation products in both alloys (Ref 11). Gold grid lines are applied to the sample surface by sputtering.

Ambient temperature tensile tests were performed at a strain rate of 3.28×10^{-5} l/s (where l is the original length) to a total strain of 3%. Creep tests were then performed at 95% of the measured 0.2% yield stress (YS). Optical microscopy and scanning electron microscopy (SEM) micrographs were taken in the same area of the polished gage sections before and after testing to record deformation products, including interphase interface sliding, stress-induced plate formation, and slip (Ref 11).

Following testing, transmission electron microscopy (TEM) specimens were sectioned from the gage length (deformed material) and from the grip section (undeformed material) using a diamond wafering blade. Disks that were 3 mm in diameter were punched from each slice, and were then thinned to less than 100 μm by tripod polishing using 800-grit paper and depth monitoring (Ref 15). The specimens were then dimpled on both sides using 3 μm diamond abrasive paste to an approximate thickness of 20 μm. Final thinning was then performed by ion milling 15° incident to the sample surface. A transmission electron microscope, operating at 300 keV, was used to obtain the required diffraction patterns and images for this study.

3. Results

The tensile and creep deformation mechanisms of the single-phase β alloys of the Ti-Mn and Ti-V systems have been explored during the course of previous studies (Ref 16-21). The deformation mechanisms of the single-phase β alloys, Ti-13.0wt.%Mn and Ti-14.8wt.%V, are dependent on grain size as well as their stability. The stability of the β alloys is given by molybdenum equivalency (MoE) (Ref 22). It is a relative measure of the stability of a β titanium alloy with any number of stabilizing elements compared with a Ti-Mo alloy with equivalent stability. The MoE has been calculated as 19.9 for Ti-13.0wt.%Mn and 9.9 for Ti-wt.%14.8V (Ref 23); therefore, the stability of Ti-13.0wt.%Mn is much higher than that of Ti-wt.%14.8V. The Ti-13.0wt.%Mn alloy has a YS of 940 MPa, and deforms solely by coarse and wavy slip, even in specimens with relatively large grain sizes (200 μm) (Ref 11). In contrast, the Ti-14.8wt.%V alloy with large grains (350 μm) has a YS of 774 MPa, and deforms by slip and the formation of stress-induced $\{332\}\langle 113\rangle$ twins (Ref 18, 24). These twins contained two orientations of the ω phase. Specimens with smaller grain sizes (18–35 μm) had YS values ranging from 876 to 900 MPa and deformed solely by slip (Ref 18).

The two-phase Ti-6.0Mn and Ti-8.1V alloys were tensile tested at ambient temperature at a strain rate of 3.28×10^{-5} l/s to a total strain of 3% (Ref 11). The Ti-6.0wt.%Mn alloy exhibits a slightly higher YS than the Ti-8.1wt.%V alloy, while the elastic modulus of each alloy is approximately equal. The Ti-6.0wt.%Mn alloy yields at 623 MPa and shows strain hardening, while the Ti-8.1wt.%V alloy yields at 597 MPa and exhibits no significant strain hardening (Ref 11). The creep strain of the Ti-6.0wt.%Mn alloy is lower than that of the Ti-8.1wt.%V alloy, which is attributed to a difference in β phase deformation mechanisms.

The deformation of Ti-6.0Mn is due primarily to fine slip in the α phase, as well as limited α phase twin formation and interphase interface sliding. The SEM observations of the sample surface following testing revealed a small number of stress-induced plates in the α phase. A TEM analysis of the Ti-6.0wt.%Mn alloy following tensile and creep testing revealed extensive slip, as well as infrequent twins, in the α phase. Twins were identified as $\{10\bar{1}1\}$ type. Slip occurred primarily with screw dislocations on prism planes with $\vec{b} = 1/3\langle 11\bar{2}0\rangle$. No slip or stress-induced plates were observed in the β phase of the Ti-6.0wt.%Mn alloy.

In contrast to the Ti-6.0wt.%Mn alloy, the SEM analysis of Ti-8.1wt.%V alloy revealed coarse deformation products spanning many α and β grains in the tensile specimens. These are absent in the Ti-6.0wt.%Mn specimens. These coarse deformation features were aligned both parallel to and across the length of the α-β interfaces, depending on the specimen area. The difference in deformation mechanisms is evident in the SEM micrographs of the gage sections of Ti-6.0wt.%Mn and Ti-8.1wt.%V following tensile testing, as shown in Fig. 1 (Ref 11).

A TEM analysis of the Ti-8.1wt.%V specimens revealed that the coarse lines visible in SEM micrographs are actually a

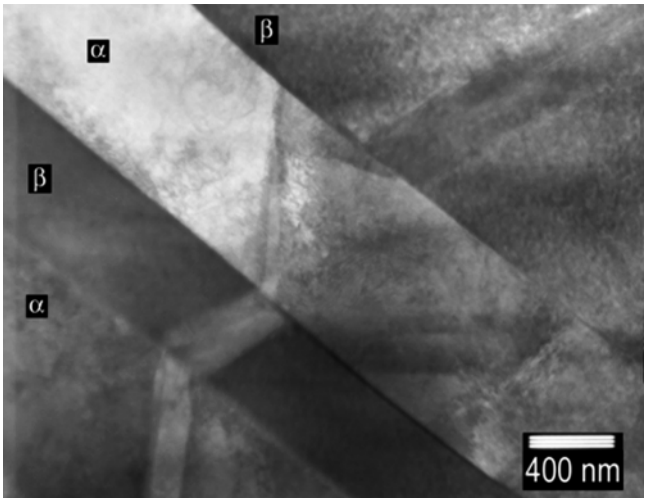

Fig. 2 TEM micrograph of Ti-8.1wt.%V alloy following tensile deformation showing stress-induced hexagonal martensite plates in the β phase and $\{10\bar{1}1\}$ twins in the α phase, alternating over large numbers of α and β grains

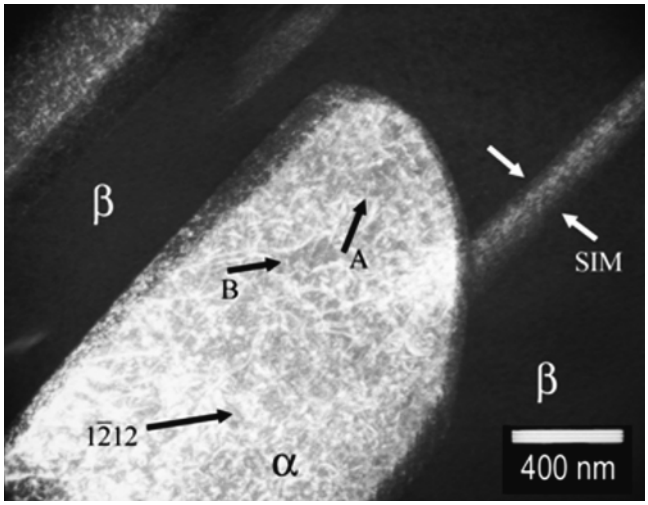

Fig. 3 Dark-field image of the α plate from Fig. 6. Screw dislocations are visible on both slip system $A = (\bar{1}100)\pm1/3[11\bar{2}0]$ and slip system $B = (10\bar{1}0)\pm1/3[1\bar{2}10]$. Arrows indicate the projection of Burgers vector. The stress-induced martensite (SIM) plate is indicated in the β phase.

combination of stress-induced hexagonal martensite plates (α′) in the β phase and either $\{10\bar{1}1\}$-type twins or coarse slip in the α phase. This is the first time twinning in the α phase was reported in association with stress-induced martensite in the β phase of a two-phase titanium alloy. It was often observed that plates of martensite in β phase alternated with twins or coarse slip in the α phase over a distance of many α and β platelets.

Numerous micrographs and diffraction patterns were taken to confirm that the stress-induced plates in the β phase were indeed hexagonal martensite. The TEM bright-field micrograph shown in Fig. 2 shows a series of martensite plates in the β phase and $\{10\bar{1}1\}$-type twins in the α phase. A dark-field image showing slip in the α phase accompanying a martensite plate in the β phase is shown in Fig. 3. Screw dislocations on prism planes are present throughout this α grain. Prism slip with "a" type Burgers vector and screw orientation is the predominant slip system in the Ti-8.1wt.%V alloy. Figure 4 shows two stress-induced martensite plates in the β phase, both with a Burgers orientation relationship with the β phase and a $\{10\bar{1}1\}$ twin relationship with the α phase. These results clearly show that the β phase deformation mechanisms in a two-phase alloy can differ from those of an equivalent single-phase β alloy.

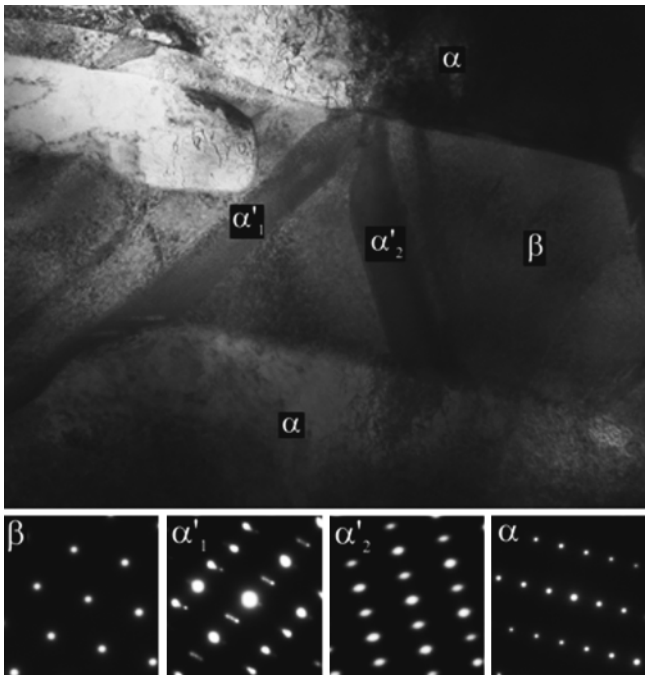

Fig. 4 Bright-field TEM micrograph showing two martensite plates in the β phase. The accompanying selected area diffraction patterns show the Burgers orientation relationship of the martensite plates to the β phase and a $\{10\bar{1}1\}$ twin relationship to the adjacent α phase.

4. Discussion

Two phenomena are of interest when examining the β phase deformation mechanisms of the Ti-V and Ti-Mn systems. The first is deformation by stress-induced martensite in the β phase of the two-phase alloy Ti-8.1wt.%V, when deformation in the matching single-phase β alloy, Ti-14.8wt.%V, occurs by twinning. The second is the formation of stress-induced martensite in the β phase of Ti-8.1wt.%V but not in the β phase of Ti-6.0wt.%Mn. In the above cases, the Burgers orientation relationship illustrated in Fig. 5 plays an important role in the interactions between the α and β phases in alloys with a Widmanstätten microstructure. An explanation of the influence of the β phase on the deformation behavior of the α phase, as well as the influence of α phase on the deformation of the β phase, is presented below.

4.1 Differing β Phase Deformation Mechanisms Between Single-Phase and Two-Phase Ti-V Alloys

The first phenomenon of interest is the formation of stress-induced hexagonal martensite plates in the β phase of Ti-8.1wt.%V in contrast to the formation of {332}-type twins in the same β phase when present as a single phase (Ti-14.8wt.%V). Two contributions explain the difference in the deformation mechanisms between the single- and two-phase

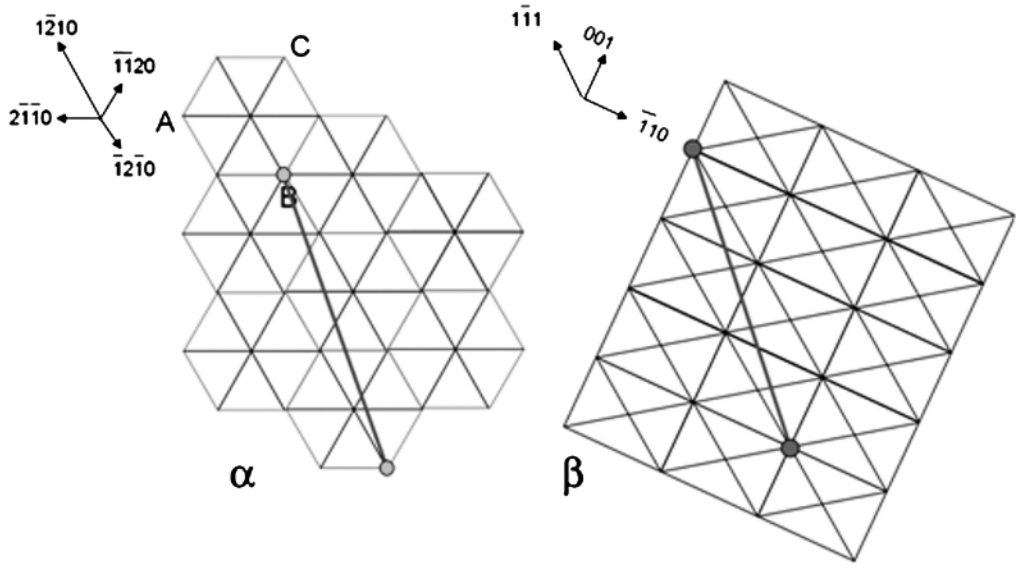

Fig. 5 Illustration of the $[1\bar{2}10]_\alpha$ $(0001)_\alpha$//$[1\bar{1}1]_\beta(1\bar{1}1)_\beta$ Burgers orientation relationship between the α and β phases in an α-β titanium alloy with Widmanstätten microstructure. The interface planes are $(\bar{5}140)_\alpha$//$(\bar{3}34)_\beta$, which are normal to the $(0001)_\alpha$ and $(110)_\beta$ planes, and are indicated as traces. The figure is drawn to scale.

alloy, α phase templating and resolved shear stresses from α phase deformation products.

4.1.1 Elastic Interaction Stresses. Elastic interaction stresses between the α and β phases in titanium alloys may be partially responsible for stress-induced martensite formation in the β phase. The elastic interaction stresses were analyzed for two-phase titanium systems in several studies (Ref 7, 8, 25). Ankem and Margolin (Ref 7, 8) analyzed the interaction stresses in two-phase titanium alloys with Widmanstätten microstructure for numerous parallel α and β slip systems. The yield strength of the α phase is lower than that of the β phase, and the elastic modulus of the β phase is lower than that of the α phase. The same situation exists for the α and β phases of both Ti-6.0wt.%Mn and Ti-8.1wt.%V.

The well-defined interface between the phases constrains the α and β phases, such that the strain on each phase must be equal. Initially, at a given applied stress on the two-phase alloy, the stress and strain of the α phase will be increased in excess of the applied stress level, while the stress on the β phase will be decreased to maintain an equivalent strain on each phase at the interface. Therefore, the α phase can deform plastically at applied stress levels that would normally result only in elastic deformation in a single-phase alloy.

After significant plastic deformation occurs in the α phase, and the strain on the alloy extends to the right of the intersection of the α and β stress-strain plots, interaction stresses are now placed on the β phase. The β phase must now constrain the plastically deforming α phase, which increases the stress in the β phase while lowering the stress on α. The additional stress acting on the β phase due to elastic interaction with the α phase contributes to β phase deformation, including the formation of stress-induced martensite in the β phase of Ti-8.1wt.%V.

4.1.2 α Phase as a Template for Stress-Induced Martensite. The activation energy for stress-induced martensite may be lowered in a two-phase titanium alloy by the α phase acting as a template for the formation of stress-induced martensite. The well-defined orientation relationship between the α and β phases, and between the β phase and hexagonal martensite, leads to a $\{10\bar{1}1\}$ twin relationship between the α phase and a martensite plate within the β phase. The TEM-selected area diffraction patterns confirmed that a martensite plate and an adjacent α grain share a common $\{10\bar{1}1\}$-type plane. This orientation relationship is shown in Fig. 5. There is no such relationship between the α phase and a {332}-type twin in β phase. The common plane between the α phase and hexagonal martensite lowers the nucleation energy of martensite formation compared with the activation energy required for twinning in a two-phase alloy. Because this nucleation template does not exist in the single-phase β alloy with identical chemistry and stability, twins are the preferred deformation product.

4.1.3 Alignment of α and β Slip and Shear Systems. To facilitate the stress-induced martensitic transformation by overcoming the activation energy, additional stresses can be added to the applied stress. The strain field resulting from dislocations, and presumably twins, can contribute to the stress required to nucleate stress-induced martensite (Ref 26). The additional energy provided by these deformation features lowers the nucleation energy barrier for martensite formation.

The resolved shear stresses from slip or twinning in the α phase that act on the shear systems for either twinning or stress-induced martensite in the β phase were calculated for all possible combinations of observed slip and twins in the α phase, and for martensite and twins in the β phase. The shear systems for twins in the β phase of the Ti-V system are of the type $\{332\}\langle 113\rangle$. In the case of stress-induced hexagonal martensite, Otte (Ref 9) proposed that the shear systems involved in the β to α′ transformation are:

$$[111]_\beta(11\bar{2})_\beta \equiv [2\bar{1}\bar{1}3]_{\alpha'}(\bar{2}112)_{\alpha'}$$

and

$$[111]_\beta(\bar{1}01)_\beta \equiv [2\bar{1}\bar{1}3]_{\alpha'}(\bar{1}011)_{\alpha'}$$

Deformation products in the α phase are predominately a-type prism slip with Burgers vectors $\langle 11\bar{2}0\rangle$ and $\{10\bar{1}1\}$ twins. Ad-

ditionally, the resolved stress from a-type basal slip was calculated due to the possibility of basal slip within the α phase in titanium alloys.

The resolved shear stress from prism slip in the α phase indicates that in general the slip systems in α phase act strongly on both the twin and martensite shear systems in the β phase. The calculations for basal slip in α phase show a favorable amount of resolved shear stress from an α slip system (>85%) on several martensite shear systems.

Calculations of the shear stress from $\{10\bar{1}1\}$-type twins in the α phase acting on the shear systems of twinning and stress-induced martensite in the β phases indicate that shear stress from twins in the α phase generally resolve with greater magnitude onto martensite shear systems than twin shear systems in the β phase. A large number of shear systems resolve more than 90% of their shear stress on several martensite shear systems.

The above conditions are only possible in a two-phase alloy. In the single-phase alloy, there is no additional stress from twinning in the α phase to act on the shear systems of martensite in the β phase. Therefore, twinning is the predominant deformation mechanism in the single-phase Ti-14.8wt.%V β alloy rather than stress-induced martensite.

4.2 Stress-Induced Martensite in Ti-8.1wt.%V But Not Ti-6.0wt.%Mn

The second phenomenon of interest is the formation of stress-induced martensite in the β phase of Ti-8.1wt.%V, but not in the β phase of Ti-6.0wt.%Mn. The α phases of these two alloys are similar, and also deform by similar mechanisms, whereas deformation in the β phase is drastically different. The reasons for different β deformation behavior include differences in the strength ratio of the β to α phases as well as the β phase stability and the presence of ω phase. The effect of these factors will be discussed in detail below.

4.2.1 β-to-α Phase Yield Strength Ratio. In Ti-6.0Mn, the YS ratio of the β to α phase is ~4, whereas the Ti-8.1V has β-to-α YS ratio of ~2.65. The tensile stresses applied to each two-phase alloy were comparable (623 versus 597 MPa), and the α phases of each alloy deform by the same mechanisms, resulting in similar stresses from interaction and α deformation products. One consequence of a lower β-to-α stress ratio in Ti-8.1wt.%V is that the additional stress required to deform β plastically, once α has begun to deform plastically, is much less than the increase in stress required for Ti-6.0wt.%Mn. While this required stress level is achieved for the β phase of Ti-8.1wt.%V, and stress-induced martensite is the result, the stress level on the β phase of Ti-6.0wt.%Mn is insufficient for the formation of stress-induced martensite. Strain from α phase deformation is instead relieved in Ti-6.0wt.%Mn by interphase interface sliding.

4.2.2 β Phase Stability and ω Phase. The martensitic transformation occurs at a temperature M_s, at or below which the transformation will happen spontaneously. Stress-induced martensite occurs because the application of stress to the material effectively raises the transformation temperature M_s to a temperature equivalent to the test temperature (Ref 27-29). Martensitic transformation is more likely in lower stability β alloys. The β phase of the Ti-6.0wt.%Mn is more stable than that of the Ti-8.1wt.%V alloy, so the transformation to martensite is more likely in the β phase of Ti-8.1wt.%V.

The presence of the athermal ω phase is a result of the instability of the β phase. Athermal ω phase is present as nanostructured crystals within the β phase of Ti-8.1wt.%V, whereas only faint traces of ω phase are present in the Ti-6.0wt.%Mn alloy. Kuan et al. (Ref 2) suggested that the metastable ω phase is an intermediate phase between the β and α phases. It is, therefore, likely that when ω phase is present the activation energy as well as the free-energy change required to form martensite in the β phase is lowered. The more stable β phase of Ti-6.0wt.%Mn requires higher stresses to overcome the higher activation energy to form stress-induced martensite. Because this stress condition is not met, stress-induced martensite does not form, and instead interphase interface sliding relieves the strain from slip and twinning in the α phase in this alloy.

For single-phase Ti-V alloys with vanadium contents ranging from 5 to 40 wt.%, several types of deformation products have been reported, including slip, twinning, several types of stress-induced martensite, or ω phase plates (Ref 2, 18, 28, 30-32). It appears that the deformation products are related to the amount of the β-stabilizing element, and, hence, the stability, in the β phase. As the vanadium concentration, and, therefore, stability, is increased, the predominant deformation mechanisms change from stress-induced martensite to ω phase plate formation, to twinning, and finally to slip. This is most likely due to the relationship between the martensite start temperature, M_s, and the amount of the stabilizing element. As the amount of β-stabilizing element is increased, the M_s temperature decreases, which makes the formation of stress-induced martensite more difficult. At higher stabilities, the stress to deform by twinning and slip is lower than that to form stress-induced martensite. The M_s temperature of the β phase of Ti-8.1wt.%V is higher than that of the β phase of Ti-6.0wt.%Mn (Ref 33), so the formation of stress-induced martensite is favored in Ti-8.1V. The presence of the α phase in effect raises the M_s temperature even further, which lowers the required energy for the nucleation of stress-induced martensite. This nucleation energy falls below that for twinning, hence, martensite is the β phase deformation mechanism in the two-phase alloy Ti-8.1wt.%V, whereas twinning occurs in the single-phase alloy Ti-14.8wt.%V.

5. Conclusions

These results are applicable to any two-phase alloy with a Burgers orientation relationship between phases. Unexpected deformation mechanisms may be the result of elastic interaction effects, deformation products in an adjacent phase, differences in the strength ratio between phases, and the stability of either phase. The specific conclusions from this study are:

- Stress-induced hexagonal martensite forms in the β phase of the α + β alloy Ti-8.1wt.%V. In contrast, twinning and slip are the deformation mechanisms in the single-phase β alloy, Ti-14.8wt.%V.
- Stress-induced martensite formation in the β phase of Ti-8.1wt.%V is due to elastic interaction stresses, with the α phase acting as a template, and the resolved shear stresses from slip and twinning in the α phase on the shear systems for stress-induced martensite in the β phase.
- Tensile and creep deformation mechanisms are limited to slip and twinning in the α phase of Ti-6.0wt.%Mn, with no

significant deformation products in the β phase, whereas the deformation products of Ti-8.1wt.%V included slip and twinning in the α phase, and stress-induced hexagonal martensite (α′) in the β phase. The difference in β phase deformation mechanism is attributed to differences in the β-to-α strength ratio and β phase stability.

Acknowledgments

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