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INFLUENCE OF THE ω PHASE ON THE TENSILE AND CREEP DEFORMATION MECHANISMS OF α-β TITANIUM ALLOYS

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Abstract. In this investigation the role of the nanocrystalline ω phase on the formation of ω' in the bulk two-phase titanium alloys with Widmanstatten microstructures is examined. The α-ω misfit strain places stress on ω planes that assist the ω → α' transformation, and shear stresses from slip and twinning in the α phase act on the ω → α' transformation shear systems. TEM and HREM were used to measure the α and ω lattice parameters, determine the orientation relationships of α, β, ω, and α' and observe the nature of the α-ω and α-α' interfaces.

1. INTRODUCTION

It was observed recently that the β phase of the two-phase alloy Ti-8.1V deforms in tensile and creep tests by stress induced hexagonal martensite (α'), while the single phase β alloy with the same composition, Ti-14.8V, deforms by slip and {332}<113> twinning [1]. The difference in deformation mechanisms has been attributed to the presence of the α phase. Influences of the α phase include elasto-plastic interactions, the α phase acting as a template for hexagonal martensite, and resolved shear stresses from α phase slip and twinning systems acting on shear systems for the β → α' transformation. In this investigation, consideration is given to the effect of the interactions between the α phase and the nanocrystalline ω phase, which exists within the β phase, on the β phase deformation mechanisms.

In β titanium alloys the formation of the ω phase is dependent upon the alloy stability. In Ti-V alloys, athermal ω phase will form in alloys with V contents up to about 20 wt.% V upon quenching [2,3]. The vanadium content of the β phase of Ti-8.1V is 14.8 wt.% V. Therefore significant ω phase is expected, and is observed in selected area diffraction patterns and dark field images of the β phase of Ti-8.1V.

The α phase has a hexagonal closed packed (HCP) structure, and the β phase has a body centered cubic (BCC) structure. The ω phase, which exists as a nanostructured phase within the beta phase of Ti-8.1V, has a P6/mmm hexagonal structure (Strukturbericht C32 designation) [4-6]. Although the α and ω phases are both hexagonal, their crystal structures are quite different, α having a c/a ratio of 1.60 and ω having a c/a ratio of 0.61.

There is a well defined orientation relationship between the α and β phase in titanium alloys with a Burgers orientation relationship. When the ω phase is present, there is an orientation relationship between four different orientations of the ω phase and

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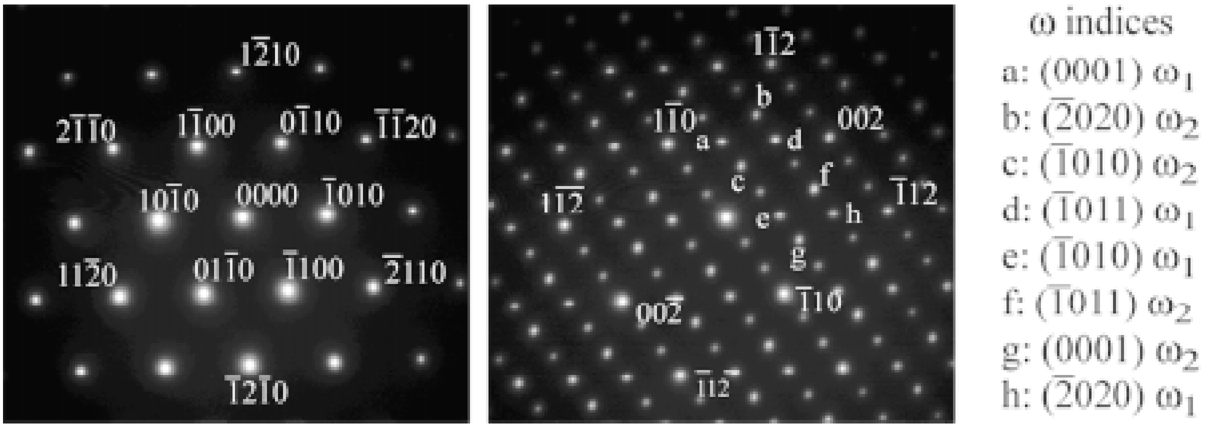

Fig. 1. Selected area diffraction patterns of the (a) α phase oriented along $[0001]_\alpha$ and the (b) β phase oriented along the $[110]_\beta$ zone axis. The diffraction patterns of ω_1 and ω_2 oriented along $[\bar{1}2\bar{1}0]_{\omega1}$ and $[1\bar{2}10]_{\omega2}$ are contained within the β phase pattern. The above diffraction patterns were taken from adjacent grains without tilting the specimen.

the β phase [7]. The orientation relationship between the α and β phases is:

α: $(0001)[1\bar{2}10]_\alpha \, // \, (110)[1\bar{1}1]_\beta$

Four variants of the ω phase have orientation relationships with the β phase:

ω_2: $(\bar{1}010)[0001]_{\omega1} \, // \, (\bar{1}12)[1\bar{1}1]_\beta$

ω_2: $(\bar{1}010)[0001]_{\omega2} \, // \, (1\bar{1}2)[\bar{1}11]_\beta$

ω_3: $(\bar{1}010)[0001]_{\omega3} \, // \, (\bar{1}\bar{1}2)[111]_\beta$

ω_4: $(\bar{1}010)[0001]_{\omega4} \, // \, (112)[\bar{1}\bar{1}1]_\beta$

These relationships have been confirmed by selected area diffraction from the $[0001]_\alpha//[110]_\beta$ zone axes of Ti-8.1V with Widmanstätten and equiaxed microstructures. The diffraction patterns for two orientations of the ω phase are visible in any given diffraction pattern from the <110> β zone axis. The $[110]_\beta$ pattern contains patterns from the $[\bar{1}2\bar{1}0]_{\omega1}$ and $[1\bar{2}10]_{\omega2}$ zone axes. These diffraction patterns are presented above in Fig. 1. The orientation rela-

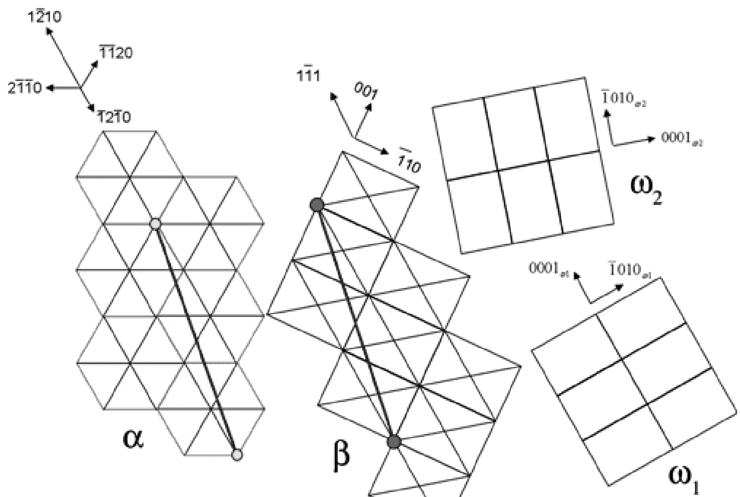

Fig. 2. Projections of the $(0001)_\alpha$, $(0001)_\beta$, and $(\bar{1}2\bar{1}0)_{\omega2}$ and $(1\bar{2}10)_{\omega2}$ planes showing the orientation relationship of the α, β, and ω phases. Plane traces are perpendicular to the projection plane, and are drawn as lines, with line intersections indicating atomic positions. The blue lines are traces of the α/β interface planes. Illustration is to scale.

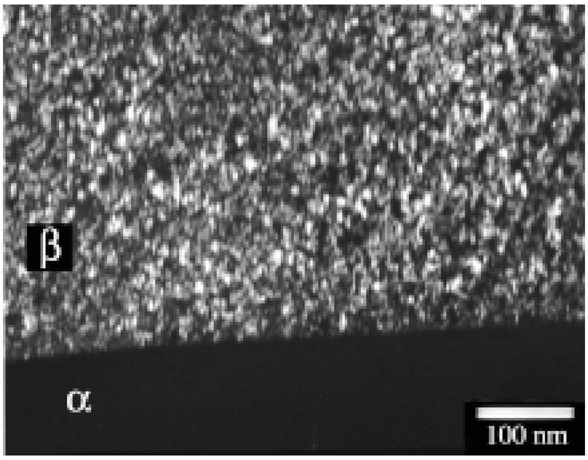

Fig. 3. Dark field TEM micrograph of Ti-8.1wt.%V alloy. Bright spots are nanocrystalline ω phase particles within the β phase.

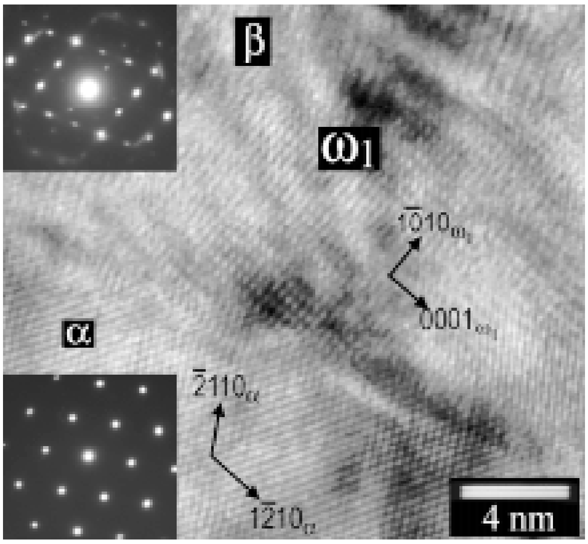

Fig. 4. HREM image of the α-ω-β interface of Ti-8.1V with accompanying selected area diffraction patterns. The ω phase shown is orientation 1. Notice the lattice distortion of the α and ω phases at the interface. Inset selected area diffraction patterns show the orientation of the specimen to the electron beam of $[0001]_\alpha//[110]_\beta//[\bar{1}2\bar{1}0]_{\omega 1}$.

tionships of α, β, and ω_1 and ω_2 are shown schematically in Fig. 2.

Due to these well defined orientation relationships, interactions can occur between the α, β and ω phases. The present work focuses on how α/ω and β/ω interactions affect the β phase deformation mechanisms in α-β titanium alloys. Although Ti-8.1V is used as the model system in this work, these results are applicable to any two-phase titanium alloy with a Burgers orientation relationship between the α and β phases when ω phase is present within the β phase.

2. EXPERIMENTAL

2.1. TEM and HREM

TEM specimens were sectioned from the gage length (deformed material) and from the grip section (undeformed material) using a diamond wafering blade. Disks 3mm in diameter were punched from each slice, which were then thinned to less then 100 μm by tripod polishing using 800 grit paper and depth monitoring [8]. 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 JEOL JEM-4000FX transmission electron microscope, operating at 300 KeV, was used to obtain the diffraction patterns and images for this study.

In order to view the ω phase and ensure that the HREM images were taken without any overlap of the α and β phases the sample was tilted along the $(0001)_\alpha$ and $(110)_\beta$ zone axes. As shown in Figs. 1 and 2, the α-β interface plane is parallel to the beam direction in this specimen orientation, and two orientations of the ω phase are visible. In order to reveal the α/β/α' orientation relationship, the specimen was positioned with the beam parallel to $[1\bar{1}1]_\beta//[1\bar{2}10]_\alpha$. In this orientation the beam is parallel to the α/α' interface plane, and the diffraction patterns of both the α phase and α' are clearly visible. High resolution microscopy of the α/α' twinned interface can only be performed in areas where the interface lies in the correct orientation very close to the foil edge, which is extremely rare and difficult to locate.

2.2. Calculation of resolved shear stresses from α phase deformation products

The α and ω phase coordinate systems were converted to the β phase coordinate system using the orientation relationships above and a series of transformation matrices. The directions and planes important for slip and twinning in α and for the martensitic transformation in ω were then converted to equivalent vectors in the β phase coordinate sys-

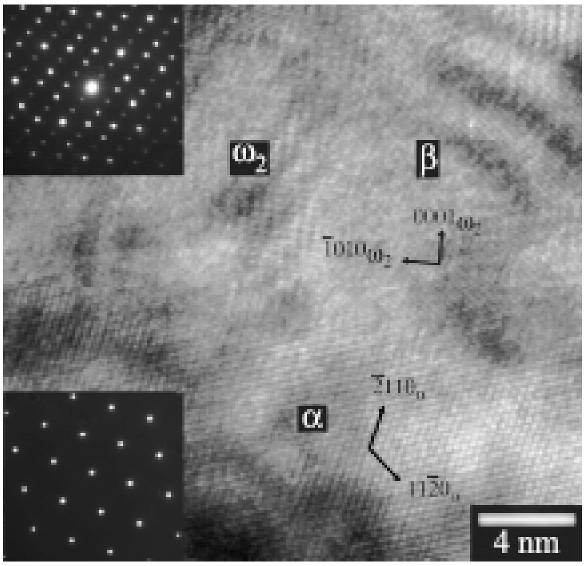

Fig. 5. HREM of the α–ω–β interface of Ti-8.1V with accompanying selected area diffraction patterns. The α phase is nearly in contact with a region of ω_2. Inset selected area diffraction patterns show the orientation of the specimen to the electron beam of $[0001]_\alpha//[110]_\beta//[1\bar{2}10]_{\omega 2}$.

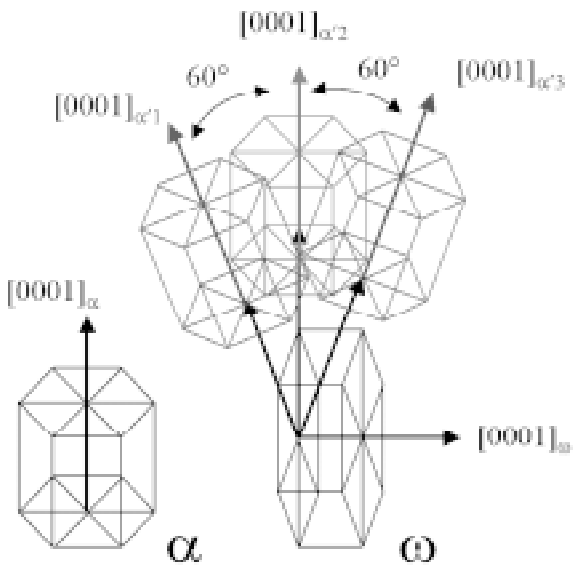

Fig. 6. Schematic of the ω to hexagonal martensite transformation, which can result in three different orientations of martensite for each orientation of ω. Because the $\{0001\}_{\alpha'}$ planes of martensite form parallel to $\{1\bar{2}10\}_\omega$ planes [12], one orientation of martensite will have the same orientation as the α phase, and the other two will be rotated 60° with respect to the α phase.

tem. Calculation of the resolved shear stresses from slip and twinning in the α phase on shear systems in the ω phase was then possible using this common coordinate system. The β phase coordinate system is convenient not only because of the orientation relationships with the α and ω phases, but also because calculations are easily performed in the cubic system. Calculation of the resolved shear stresses from 'a' type (b = 1/3<$11\bar{2}0$>) prism and basal slip and $\{10\bar{1}1\}$ and $\{10\bar{1}2\}$ type twins in the α phase on the ω to α' shear systems was performed by tensor transformation.

3. RESULTS AND DISCUSSION

3.1. α-ω interface

The ω phase is present in high concentrations in the β phase of Ti-8.1V. The exact concentration is difficult to determine due to the multiple orientations and the small particle size in relation to the sample thickness. A dark field image of nanocrystalline athermal ω phase within the β–phase of Ti-8.1V using a single ω reflection is shown in Fig. 3. Only one of four orientations of the ω phase is shown when imaged in this manner. The bright areas are ω phase.

Due to the high concentration of ω phase, it seems reasonable that along the α-β interface there will be a great number of α-ω interfaces. This was confirmed with HREM images of the interface. Orientations 1 and 2 of the ω phase were observed in proximity to the α phase. Fig. 4 shows the interface of the α phase with a region of ω_1, and Fig. 5 shows the α phase nearly in contact with ω_2.

3.2. ω → hexagonal martensite (α') transformation

The α to ω phase transformation is a reversible martensitic transformation, where α is transformed to ω under hydrostatic stress [4,9-11]. Therefore, the α → ω and ω → α transformation pathways are equivalent. According to the work of Silcock [12], the α phase is transformed to the ω phase by a series of shears of $(1\bar{1}00)_\alpha$ planes in the $[11\bar{2}0]_\alpha$ direction. This results in a transformation where the

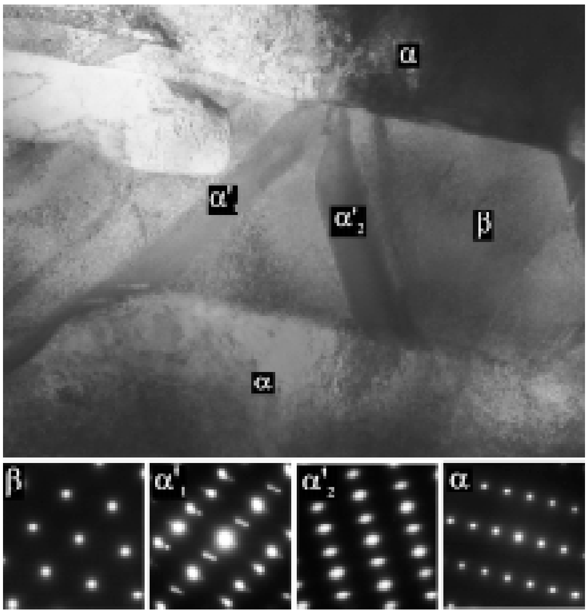

Fig. 7. Bright field TEM micrograph showing two stress-induced hexagonal martensite plates (α') within the β phase of Ti-8.1V tested in tension. Accompanying diffraction patterns show that both martensite planes have a $\{10\bar{1}1\}$ near twin relationship to the α phase, and that the martensite forms with $\{0001\}_{\alpha'}$ planes parallel to $\{110\}_\beta//\{11\bar{2}0\}_\omega$ planes.

previous $(0001)_\alpha$ plane is now a $(11\bar{2}0)_\omega$ plane. The reverse transformation takes place by shear on the $(1\bar{1}00)_\omega$ planes in the $[11\bar{2}0]_\omega$ direction, resulting in $(0001)_\omega$ planes parallel to the previous $(11\bar{2}0)_\omega$ planes. In this way each ω crystal can transform into three orientations of α with the c axis oriented 60° from one another.

There are other transformation pathways presented for the α-ω transformation [4,10,11], but these are not relevant for ω phase that is contained within the β phase. Instead, these studies are concerned with ω phase that is transformed in HCP metals (Ti and Zr) under high hydrostatic pressures at elevated temperatures, resulting in bulk ω or ω within α. In such a case the transformations would not maintain a Burgers orientation relationship between the transformed α' and the surrounding β phase, which has been observed in Ti-8.1V and is discussed below. Therefore the shear transformation proposed by Silcock is used for this analysis.

3.3. Resulting α/α' orientation

The four orientations of ω phase are each aligned such that their basal ($\{0001\}_\omega$) planes are parallel to

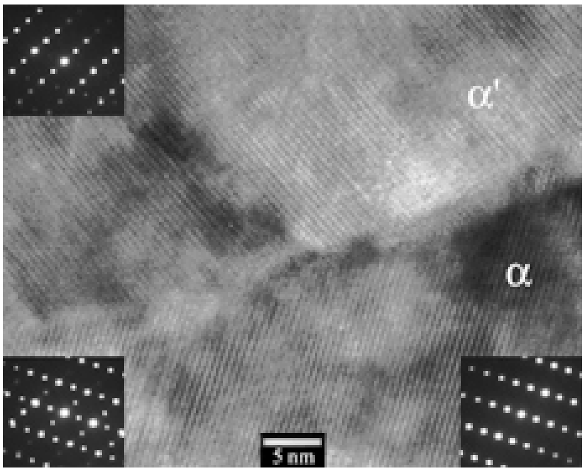

Fig. 8. High resolution TEM micrograph showing the interface of stress induced martensite, α', and the α phase. Notice the $\{10\bar{1}1\}$ near twin relationship between the phases. Selected area diffraction patterns from each phase and the interface taken parallel to the $(1\bar{2}10)$ zone axis are inset.

the $\{111\}_\beta$ type planes in the β phase, and $\{11\bar{2}0\}_\omega$ planes are parallel to $\{110\}_\beta$ planes. During the transformation from ω to α' the $\{11\bar{2}0\}_\omega$ planes transform to $\{0001\}_{\alpha'}$ [9,12]. This means that any one of the four orientations of ω can transform into three independent orientations of α'. The resulting orientations of this transformation are illustrated schematically in Fig. 6. Due to the $\{11\bar{2}0\}_\omega//\{110\}_\beta$ orientation of ω to the β phase, all of these α' plates obey the Burgers orientation relationship with the β phase. Martensite that results from the ω_1 and ω_2 orientations of ω will either have the same orientation as the α phase, $(0001)_{\alpha'}//(0001)_\alpha$ or $(0001)_{\alpha'}$ will be rotated 60° with respect to $(0001)_\alpha$. In martensite which results from the ω_3 and ω_4 orientations, $(0001)_{\alpha'}$ will be rotated 90° or 60° with respect to $(0001)_\alpha$. The 60° rotation is nearly equivalent to a $\{10\bar{1}1\}$ twin relationship with α, which has a rotation between basal planes calculated as 58° in Ti-8.1V. This angle may change slightly in other alloys due to differences in the c/a ratio. The 90° rotation would be close to the 94° rotation of a $\{10\bar{1}2\}$ twin. The $\{10\bar{1}1\}$ near-twin relationship has been observed between stress induced martensite and the α phase of Ti-8.1V, and is shown by the selected area diffraction patterns in Fig. 7. Fig. 7 is also an example of martensite plates forming from different orientations of omega phase within a single β grain, where each plate has a twinned relationship to the α phase, and both martensite plates will obey the Burgers orientation relationship with the β phase. Therefore, the interface created between the α phase and a stress induced hexagonal martensite plate should be coherent, with either a matching or twin-like interface. The coherency of the twin-like interface was confirmed using HREM, and is shown in Fig. 8. This is in contrast to the α-ω interface that exists prior to the $\omega \rightarrow \alpha'$ transformation, which is discussed below.

3.4. α-ω misfit strain

As discussed above, the α'/α interface is coherent. This is in contrast to the observed misfit at the α/ω interface. Both the α and ω phases at the interface appear distorted due to misfit strain.

The misfit between the α and ω_1 and ω_2 phases has been calculated for orientations 1 and 2 of ω along the α/ω interface parallel to the prior α-β interface. This interface is important because of the nature of α and ω formation in the Widmanstatten alloys. When the alloy is cooled below the β transus temperature, the α phase grows from the β phase, resulting in a fixed α/β interface plane of $(\bar{5}140)_\alpha$/ $/(\bar{3}34)_\beta$ given by the Burgers orientation relationship described above and illustrated in Fig. 2. The Burgers orientation relationship was also observed in Ti-8.1V with an equiaxed microstructure, so this analysis applies to both microstructures. When the alloys are quenched from 963K, athermal ω phase nucleates throughout the β phase. When the ω phase forms at the α/β interface the interface plane remains unchanged, resulting in an α/ω interface plane that is parallel to the prior α/β interface. These planes have been calculated as $(20\bar{2}7)_{\omega 1}$ and $(\bar{1}0112)_{\omega 2}$ for ω_1 and ω_2 respectively.

Strain has been calculated along the $[0001]_\alpha$ and $[1\bar{3}20]_\alpha$ directions, which are perpendicular directions in the α/ω interface plane. Strain in the direction parallel to $[0001]_\alpha$ in ω_1 or ω_2 is given by Eq. (1):

$$\varepsilon_{\omega 1,\omega 2} = \frac{c_\alpha - a_\omega}{a_\omega}. \qquad (1)$$

This strain is minimal for both ω_1 and ω_2 due to the similarly of the $\{110\}_\beta$ and $\{11\bar{2}0\}_\omega$ planes. The misfit strain between near-parallel α and ω planes in a direction parallel to the $[1\bar{3}20]_\alpha$ direction are greater than those in the $[0001]_\alpha$ direction. The best matching planes are $(0\bar{1}10)_\alpha/(0001)_{\omega 1}$ and $(1\bar{2}10)_\alpha/(\bar{1}010)_{\omega 2}$. The strains in the $[1\bar{3}20]_\alpha$ direction are given by Eqs. (2) and (3):

$$\varepsilon_{\omega 1} = \frac{\dfrac{2\sqrt{21}}{9}a_\alpha - \dfrac{\sqrt{(7c_\varpi)^2 + \left(\dfrac{\sqrt{3}}{2}a_\omega\right)^2}}{7}}{\dfrac{\sqrt{(7c_\varpi)^2 + \left(\dfrac{\sqrt{3}}{2}a_\omega\right)^2}}{7}}, \qquad (2)$$

$$\varepsilon_{\omega 1} = \frac{\dfrac{2\sqrt{21}}{9}a_\alpha - \dfrac{\sqrt{c_\varpi^{\,2} + \left(\dfrac{6\sqrt{3}}{2}a_\omega\right)^2}}{6}}{\dfrac{\sqrt{c_\varpi^{\,2} + \left(\dfrac{6\sqrt{3}}{2}a_\omega\right)^2}}{6}}, \qquad (3)$$

The measured lattice parameters for α and ω in Ti-8.1V using selected area diffraction patterns are presented in Table 1, and the calculated misfit strains in both directions using the above equations are presented in Table 2.

The interface strain acts to expand both ω_1 and ω_2 along their [0001] direction, by placing a tensile stress on the $(\bar{1}010)$ plane of ω_1 in the [0001] direction and a compressive stress on the (0001) plane in the $[\bar{1}010]$ direction in ω_2. Silcock [12] states that an expansion of ~4% in the $[0001]_\omega$ direction is

Table 1. Lattice Parameters for the α and ω phases.

Phase	a (nm)	c (nm)
α	0.2937	0.4699
ω	0.4688	0.2858

Table 2. Misfit strain between the α/ω_1 and α/ω_2 interfaces.

Interface	% Strain in ω - Direction $[0001]_\alpha$	$[1\bar{3}20]_\alpha$
α/ω_1	0.245	2.584
α/ω_2	0.245	-26.8190

Table 3. Maximum resolved shear stress from prism slip onto $\omega \rightarrow \alpha$ shear systems.

ω orientation	ω shear system	Prism Slip System $(1\bar{1}00)[11\bar{2}0]$	$(\bar{1}010)[1\bar{2}10]$	$(0\bar{1}10)[\bar{2}110]$
1	$(\bar{1}010)[0001]$	-0.51	1	0.49
2	$(\bar{1}010)[0001]$	0.15	0.78	0.93
3	$(0\bar{1}10)[0001]$	-0.53	0.38	-0.14
4	$(1\bar{1}10)[0001]$	0.53	-0.38	0.14

required to give coincidence between α and ω phases in the transformation. Silcock also states that a ~2% expansion is required in the $[1\bar{2}10]_\omega$ direction. The strain from the a phase in the $<1\bar{2}10>_\omega$ directions of ω_1 and ω_2 parallel to $[0001]_\alpha$ is ~0.25%, and places a tensile stress on the ω phases for this required expansion. Therefore, in addition to being a high energy interface, the interfacial misfit places stresses on the ω phase in a way that favors transformation of ω to the α, or hexagonal martensite, phase.

The interface between α and ω has a high interfacial free energy due to the high misfit strain. The free energy is lowered by the formation of stress induced hexagonal martensite, as the new interface will be a coherent twin-like interface. The strain energy reduction at the α/ω interface is no doubt a driving force in the ω to α' transformation.

The application of misfit stresses provides a driving force in the ω to α transformation, and forming a coherent twin-like interface between α and α' reduces the interfacial free energy of the strained α/ω interface. In these ways the α phase acts as a template for the formation of hexagonal martensite.

3.5. Calculations of resolved shear stress from slip, twinning in α on $\omega \rightarrow \alpha'$ shear systems

It is known that there is an alignment of slip systems in the α and β phases due to the Burgers orientation relationship [13-18]. In the current work the magnitudes of resolved shear stress have been calculated for α phase deformation mechanisms, slip and twinning, on the $\omega \rightarrow \alpha$ transformation shear systems discussed in Section 2 above.

Stress induced martensite plates in the β phase of Ti-8.1V were accompanied by either slip or twinning in the adjacent α phase. Previously, calculations have been performed to determine the resolved shear stresses from α phase slip or twinning on the shear systems for twinning or stress induced martensite formation in β. The ω phase is an intermediate phase in the β to α phase transformation [7]. It therefore stands to reason that shear stresses acting on the ω phase at the α-ω interface would nucleate martensite even more readily than the same shear stresses acting on the β phase. Contact of ω with the α phase transmits this stress directly to the $\omega \rightarrow \alpha'$ shear systems. In this investigation the resolved shear stress from prism slip, basal slip, and $\{10\bar{1}1\}$ and $\{10\bar{1}2\}$ twins in the α phase on the $\omega \rightarrow \alpha'$ shear transformation systems have been calculated.

Three independent prism slip systems and six independent twinning systems for both $\{10\bar{1}1\}$ and $\{10\bar{1}2\}$ type twins were considered for deformation products observed in the α phase. Slip was observed in Ti-8.1V with 'a' type screw dislocations, therefore Burgers vectors of $1/3<11\bar{2}0>$ were used in the calculations. Because 'a' type screw dislocations can also slip on the basal plane in titanium, the three independent basal slip systems were also included in these calculations. The shear stress from these a phase deformation mechanisms act in varying degrees on the three independent $\omega \rightarrow \alpha'$ transformation shear systems in each of the four w phase orientations. All possible combinations of resolved shear stress have been considered. The maximum resolved shear stresses on the ω phase transformation systems are presented below in Tables 3, 4, 5 and 6. Any negative values of resolved shear stress are a result of only independent shears being used in the calculations. The absolute values should be considered, as negative values can be made positive by simply considering shear in the opposite direction along the same shear plane.

The results of the above calculations have several consequences. First, it should be noted that the resolved shear stress from α phase slip and

Table 4. Maximum resolved shear stress from basal slip onto $\omega \rightarrow \alpha$ shear systems.

		Basal Slip System		
ω orientation	ω shear system	$(0001)[11\bar{2}0]$	$(0001)[1\bar{2}10]$	$(0001)[\bar{2}110]$
1	$(\bar{1}010)[0001]$	0.41	-0.87	0.42
2	$(\bar{1}010)[0001]$	-0.58	-0.29	0.85
3	$(0\bar{1}10)[0001]$	-0.08	0.67	-0.57
4	$(1\bar{1}10)[0001]$	0.25	0.48	-0.71

twins acts more strongly on the ω phase shear systems than β phase twinning systems. In the few instances where the resolved shear stress is greater on a twinning shear system, the magnitude of resolved stress on the β twinning system is low, and therefore the twinning will not occur. Therefore, stress induced martensite is likely to form due to the combination of applied stress and the resolved shear stress of slip and twinning on the ω phase.

Second, the ω orientations that will transform are different depending on whether slip or twinning is the primary deformation mechanism in the adjacent α phase. In the case of 'a' type prism or basal slip in α, ω orientations 1 or 2 are more likely to transform. For $\{10\bar{1}1\}$ or $\{10\bar{1}2\}$ twinning in α, ω orientations 3 or 4 will transform. The values of highest resolved shear stress for each α slip or twinning system are in bold to highlight this point.

Third, within each orientation there is a preferred transformation shear system. Although three possible martensite orientations are possible, it is most likely that one of these will form preferentially within any β grain. This means that multiple martensite plates within a β grain will have the same orientation relationships to one another, as well as the β and α phases, which has been observed experimentally. There are exceptions in cases where, for a given slip or twinning system there are more than one $\omega \rightarrow \alpha$' transformation system with equivalent resolved shear stress. This is the case for the $(10\bar{1}1)[10\bar{1}2]$ twin system, which acts on $(\bar{1}010)[0001]$ and $(0\bar{1}10)[0001]$ of ω_4 approximately equally and for the $(0001)[1\bar{2}10]$ basal slip system acting on $(0\bar{1}10)[0001]$ and $(\bar{1}010)[0001]$ of ω_1. Such an occurrence would explain martensite plates forming in two directions in a single β grain as shown in Fig. 7.

3.6. Martensite growth through ω-β interactions

It was shown above how α/ω phase interactions can aid in the nucleation of stress induced hexagonal martensite from the ω phase within the β phase of α-β titanium alloys with a Widmanstätten microstructure. The question as to how the martensite

Table 5. Maximum resolved shear stress from twins onto $\omega \rightarrow \alpha$ shear systems.

		Twin Slip System		
ω orientation	ω shear system	$(\bar{1}011)[10\bar{1}2]$	$(10\bar{1}1)[\bar{1}012]$	$(0\bar{1}11)[01\bar{1}2]$
1	$(1\bar{1}00)[0001]$	0.00	0.01	-0.22
2	$(1\bar{1}00)[0001]$	0.58	-0.31	0.18
3	$(\bar{1}010)[0001]$	-0.50	-0.80	-0.56
4	$(\bar{1}010)[0001]$	-0.8	-0.5	-0.89
ω orientation	ω shear system	$(01\bar{1}1)[0\bar{1}02]$	$(\bar{1}101)[1\bar{1}02]$	$(1\bar{1}01)[\bar{1}102]$
1	$(1\bar{1}00)[0001]$	0.59	0.22	-0.59
2	$(1\bar{1}00)[0001]$	-0.01	0.16	-0.57
3	$(\bar{1}010)[0001]$	-0.89	-0.40	-0.38
4	$(\bar{1}010)[0001]$	-0.56	-0.38	-0.40

Table 6. Maximum resolved shear stress from twins onto $\omega \rightarrow \alpha$ shear systems.

		Twin Slip System		
ω orientation	ω shear system	$(\bar{1}012)[10\bar{1}1]$	$(10\bar{1}2)[\bar{1}011]$	$(0\bar{1}12)[01\bar{1}1]$
1	$(\bar{1}010)[0001]$	0.01	0.01	-0.45
2	$(\bar{1}010)[0001]$	-0.32	-0.32	-0.20
3	$(\bar{1}010)[0001]$	-0.79	-0.75	-0.89
4	$(\bar{1}010)[0001]$	-0.75	-0.79	-0.84
ω orientation	ω shear system	$(01\bar{1}2)[0\bar{1}11]$	$(\bar{1}102)[1\bar{1}01]$	$(1\bar{1}02)[\bar{1}101]$
1	$(\bar{1}010)[0001]$	-0.45	0.45	0.45
2	$(\bar{1}010)[0001]$	-0.20	0.50	0.50
3	$(\bar{1}010)[0001]$	-0.84	-0.46	-0.47
4	$(\bar{1}010)[0001]$	-0.89	-0.46	-0.46

continues to propagate through the β phase must be considered. The answer lies in the orientation relationship of the ω and β phases. The orientation relationship $\{\bar{1}010\}\langle 0001\rangle_{\omega} // \{\bar{1}12\}\langle 1\bar{1}1\rangle_{\beta}$ means that each shear system for a $\omega \rightarrow \alpha'$ transformation is exactly parallel to a $\beta \rightarrow \alpha'$ shear system as given by Otte [18]. The transformed α' from the β phase will have the same orientation as α' from the ω phase. Therefore, once the transformation is initiated in the ω phase, it can continue through the β phase along parallel shear systems.

3.7. ω phase in the single-phase β alloy Ti-14.8V

The role of the ω phase in the formation of stress induced hexagonal martensite in two phase titanium alloys, particularly Ti-8.1V, has been discussed at length above. During this transformation the ω and β phases are consumed, such that no ω or β phase remains in the martensite plate. The β and ω shear transformation systems are parallel, which explains how the β phase transforms completely along with the ω phase.

The same β phase deforms by {332}<113> twins when present as a single-phase alloy, Ti-14.8V. Within these twins the ω phase is present, with the same orientation relationship to the twinned region of β as to the parent β [19], which has been observed by others in {332}<113> [20] and {112}<111> twins [21]. In order to maintain this relationship with the twin, the ω phase may either shear along with the β phase, or it may dissolve during twinning and recrystallize along its preferred directions within the twin when twinning is complete [21]. Although it is impossible to determine which of these mechanisms is operating, the result is the same, that ω is present within the twin.

4. SUMMARY

Interactions between the α and ω phase are possible due to the high concentration of ω phase within low stability β phases, which results in some fraction of the ω phase in contact with the a phase. The misfit strain between the α and ω phases is high, and places stress on the ω phase along planes and directions which aid in the $\omega \rightarrow \alpha'$ transformation. This misfit strain is eliminated by the transformation of ω to α', which has a twin-like orientation relationship to the α phase and a coherent interface. Also, a high magnitude of shear stress from slip and twinning in the a phase resolves upon the $\omega \rightarrow \alpha'$ shear transformation systems, aiding the transformation. Once the martensite plate has nucleated, it can continue growing though the beta phase due to parallel $\omega \rightarrow \alpha'$ and $\beta \rightarrow \alpha'$ shear transformation systems. No ω phase remains within the stress induced martensite plate, which is in contrast to the single phase β alloy Ti-14.8 which deforms by twins that contained the same ω as the parent β phase.

5. CONCLUSIONS

1. The nanocrystalline ω phase plays an important role in the bulk β phase deformation mechanisms of two-phase titanium alloys.

2. Misfit strain between the α and ω phases aids in the ω → α' transformation.
3. The interfacial free energy between the α and ω phases is reduced by the ω → α' transformation. This is because the resulting stress induced martensite has a near-twin relationship and coherent interface with the α phase due to orientation relationships between the α, β, and ω phases.
4. Shear stresses from slip and twinning in the a phase resolve onto the ω → α' shear systems, which aids in the ω → α' transformation, resulting in α' plates in the β phase. This is in contrast to the single-phase β alloy, where twinning is the primary deformation mechanism.
5. Parallel ω → α' and β → α' phase transformation shear systems allow the stress induced martensite which nucleates in the ω phase to propagate through the β phase.
6. In contrast to twins in the single phase β alloy Ti-14.8V, no ω phase remains within the stress induced martensite plates in the β phase of Ti-8.1V.

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