

A Comparative Study on the Substructure Evolution and Mechanical Properties of TIMETAL® 407 and Ti-64

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Abstract

Titanium and titanium alloys are excellent candidates for aerospace applications owing to their high strength to weight ratio. Alpha/beta titanium alloys are used in nearly all sections of the aircraft, including the fuselage, landing gear, and wing. Ti-6Al-4V is the workhorse alloy of the titanium industry, comprising of nearly 60% of total titanium production. TIMETAL® 407, Ti-0.85Al-3.9V-0.25Si-0.25Fe (Ti-407) is an excellent candidate for alloy applications requiring excellent machinability and increased energy absorption. These properties are a result of the alloy's increased ductility while maintaining moderate levels of strength. In this study, the deformation mechanisms of Ti-407 have been studied at high strain rates using split-Hopkinson bar testing. Utilizing post-mortem characterization, Ti-407 has been shown to deform significantly by $\langle c+a \rangle$ slip and deformation twinning. The observation of $\langle c+a \rangle$ slip is in contrast with other studies and will be discussed further.

1. Introduction

Two-phase alpha/beta titanium alloys are used in a wide variety of aerospace, energy, and biomedical applications. These alloys exhibit high specific strength, elastic modulus, and fracture toughness. Ti-6Al-4V is the most commonly used alpha/beta titanium alloy. It is a two phase alloy with a hexagonal close-packed alpha phase and a body-centered cubic beta phase. Three typical microstructures are commonly formed: fully equiaxed alpha, fully lamellar, and a bi-modal microstructure. The alpha phase forms during cooling from the beta phase following the Burgers orientation relationship (OR): $(0001)\alpha // (101)\beta$ and $[2\bar{1}10]\alpha // [111]\beta$.

The room temperature deformation of pure alpha-titanium has already been widely studied [1–4]. Two types of Burgers vectors are possible for gliding dislocations: **a**-type ($1/3 < 11\bar{2}0 >$), which can glide on the prismatic, basal, and first-order pyramidal planes. Prismatic glide of **a**-type dislocations accounts for the main deformation mode in CP alpha-titanium and the deformation is controlled by the motion of the screw segments [1–3,5]. Williams et al. [6] have shown that **a**-type slip on the basal plane becomes increasingly important with increasing aluminum content, reaching equal parts contribution to deformation around 5-6 wt.% aluminum. While deformation through **a**-type slip accounts for the bulk of deformation experienced in alpha-titanium, it does not bring about any dimensional change along the **c**-axis. Strain in the **c** direction is achieved through $\langle c+a \rangle$ slip with Burgers vector ($1/3 < 11\bar{2}3 >$). There are also a number of twinning modes that can accommodate shape change along the **c**-axis in the alpha phase [7].

The deformation of alpha/beta titanium alloys has been studied more recently [5,8–11]. Slip transmission between the two phases has been studied by Suri et. al [11] and Savage et al. [9,10] through room temperature deformation of single alpha-beta colony crystals in Ti-5Al-2.5Sn-0.5Fe and Ti-6Al-2Sn-4Zr-2Mo-0.1Si, respectively. One of three $\langle a \rangle$ slip vectors in the alpha phase (a_1) is nearly parallel to a $\langle 111 \rangle$ slip direction in the beta phase, resulting in easy slip transmission across the alpha/beta interface.

Recently, a new alpha/beta titanium alloy Ti-1Al-4V-0.25Si-0.25Fe-0.15O (TIMETAL®407) has been developed for alloy applications requiring increased energy absorption. Ti-407 can provide considerable cost savings through ease of processing and machining, while still exceeding the ballistic impact performance of Ti-6Al-4V. The alloy achieves these properties through a combination of high ductility and moderate strength, producing a high fracture toughness alloy [12].

In the current study, a detailed investigation of the deformation mechanisms of Ti-407 will be presented. The primary focus was to elucidate the source of the enhanced fracture toughness experienced in this alloy. A thorough TEM investigation of the dislocation substructure in Ti-407 with varying microstructures is first presented. Finally, these results are compared with the deformation mechanisms of Ti-6Al-4V.

2. Experimental Procedures

In this study, deformation substructure analysis was conducted on room temperature tensile samples of Ti-407. Initial tests were performed to determine high strain rate effects and were completed using a split-Hopkinson bar (SHB) apparatus. Samples for split-Hopkinson bar testing were prepared at Titanium Metals Corporation (TIMET) U.K. and were subsequently tested at Oxford University. All samples in this study were tested in tension. The samples for SHB testing were of duplex microstructure and contained a strong rolling texture, as shown in Figure 1. It should be noted that the images in Figure 1 are transverse cross sections, meaning the tensile axis is normal to the page. Using the image processing software, MIPAR, the samples were measured to be 90% alpha, with an equiaxed grain size of 7.1 μm , and an alpha lath thickness of 360 nm. The deformation mechanisms were studied at the mesoscale using EBSD. The substructure was then probed using transmission Kikuchi diffraction (TKD) on a FEI Apreo SEM and diffraction contrast imaging on a FEI 300kV Tecnai transmission electron microscope (TEM). Lamella samples were prepared using an FEI Nova dual-beam focused ion beam.

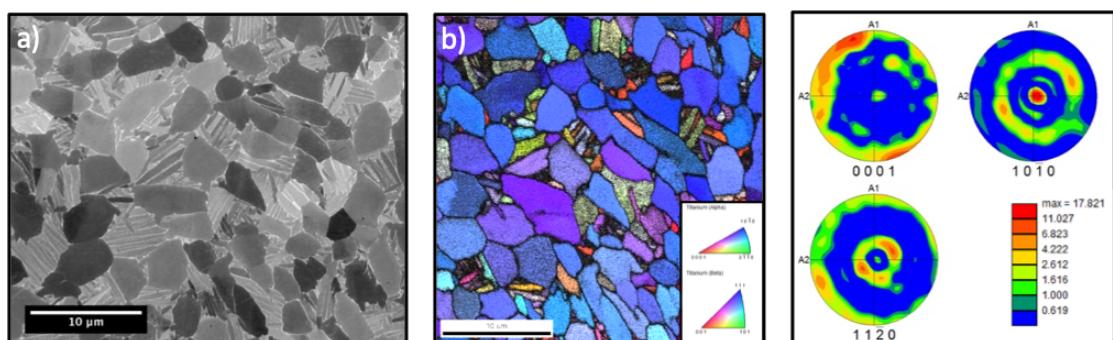


Figure 1: Microstructure of split-Hopkinson bar tests. a) backscattered electron image b) inverse pole figure showing rolling texture c) Texture plot confirming rolling texture

3. Results

TIMETAL®407 experiences high degrees of strain during deformation while maintaining moderate levels of strength [12]. This combination creates a high fracture toughness alloy. The main micromechanisms of the plastic deformation of this alloy were identified using post-mortem imaging. The features of the substructure, which were consistent for all samples tested, are first remarked upon. Figure 2 shows a typical example of the deformation substructure from split-Hopkinson bar testing.

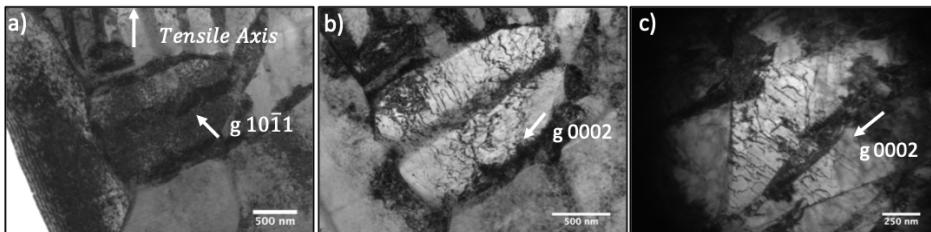


Figure 2: Diffraction contrast images of SHB sample showing a) high dislocation density in sample b) $\langle c+a \rangle$ dislocations in transformed grain and c) $\langle c+a \rangle$ dislocations in an equiaxed grain

A dense dislocation network was observed (Figure 2), which is consistent with rapid deformation and high degrees of strain experienced in split-Hopkinson bar testing. The sample also underwent non-basal deformation, evidenced by the numerous $\langle c+a \rangle$ dislocations in Figure 2b and 2c. It is clear that both equiaxed grains and transformed beta grains deform by $\langle c+a \rangle$ slip. The observation of substantial $\langle c+a \rangle$ dislocations is rather surprising due to the strong prismatic texture present in the sample (Figure 1). While the orientation of the grains in Figure 2 are not known, the likelihood of them being oriented favorably for $\langle c+a \rangle$ slip is low. If this is so, grains oriented in this manner will still have some stress resolved on their $\{10\bar{1}1\}$ planes, which is the primary slip plane for $\langle c+a \rangle$ slip [6,13,14,15], though it should be lower than the shear stress resolved on the prism plane. Deformation by $\langle c+a \rangle$ slip in this condition would thus render a non-Schmid effect. This break down of the Schmid's Law has been overserved previously in titanium [1,16,17], though primarily for a -slip. In addition to orientation, alloying content may influence deformation mechanisms. It has been shown by Williams et al. [6] and Cass et al. [18] that the propensity of $\langle c+a \rangle$ slip decreases with decreasing Al content. The extensive $\langle c+a \rangle$ slip observed in the low-aluminum containing Ti-407 is in contrast with these studies. Another important factor controlling active slip systems is the critical resolved shear stress (CRSS). The CRSS for $\langle c+a \rangle$ slip in Ti-6Al crystals and Ti-64 has been measured in tension and compression [6,14]. It was found that the CRSS for $\langle c+a \rangle$ slip in compression is higher than that in tension, though the CRSS for this slip system is always higher than the CRSS for a -slip.

In addition to $\langle c+a \rangle$ slip, deformation twinning has been observed in Ti-407, as seen in Figure 3. Note that the tensile axis is normal to the page.

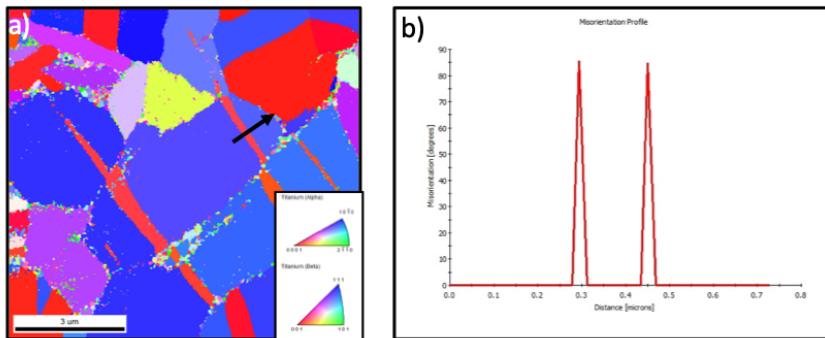


Figure 3: EBSD analysis of sample post-deformation: a) Inverse pole figure showing the presence of deformation twins and b) misorientation profile across twin

The arrow marks the direction in which a misorientation profile was measured, which shows the difference in orientation between the parent grain and twin. The difference in orientation was measured to be 86 degrees, which corresponds to a $\{10\bar{1}2\}$ type twin, and was confirmed in the TEM and shown in Figure 4a. This makes sense as this twinning mode has the lowest twinning shear. Deformation twinning is an expected deformation mechanism as the lower Al content in this alloy decreases the stacking fault energy and thus increases the propensity for deformation twinning. Notice the $\langle c+a \rangle$ slip surrounding the twin in the parent grain. Accommodation like this has been observed and modeled before [6,19,20] and most likely assists in the strain associated with the shape change experienced during twinning. The orientation of the parent grain relative to the tensile axis was measured using transmission Kikuchi diffraction (TKD) and can be seen in Figure 4b. While the fraction of grains deformed by twinning has not yet been measured, it was determined through low mag EBSD data that a significant number of grains deformed by this mechanism. The large fraction of grains that deformed by twinning is most likely due to the rapid strain rates experienced during split-Hopkinson bar testing [21-23].

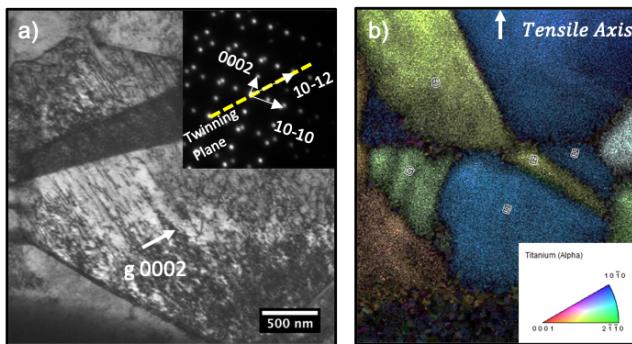


Figure 4: a) Bright field image and diffraction pattern identifying twin type and $\langle c+a \rangle$ slip accommodation and b) TKD data showing orientation of grain relative to tensile axis

While results have not been shown in this work, a systematic study on the deformation response of Ti-64, with particular focus on $\langle c+a \rangle$ slip, has been conducted and compared with Ti-407. It was observed that when the c -axis was aligned with the tensile direction, Ti-64 deformed by $\langle c+a \rangle$ slip. This is consistent with several studies [6,14,18] as this orientation promotes c -axis deformation. Deformation

twinning was never observed in any samples. When samples with a similar orientation as Figure 1 were deformed, $\langle c+a \rangle$ slip was never seen, which makes sense as grains of this orientation should deform primarily by a -slip due to the high Schmid factor and low CRSS for this slip system. This is in contrast with the findings of Ti-407, which show deformation by $\langle c+a \rangle$ slip even when displaying a strong rolling texture. This difference may account for the increased strain and fracture toughness experienced in Ti-407 compared with Ti-64.

4. Discussion

Polycrystalline Ti alloys in general show significant ductility despite the fact that they deform mostly by a -type ($1/3 < 1\bar{1}\bar{2}0 >$) slip and deformation in the c -[0001] direction is severely limited because of the differences in CRSS values between the $\langle a \rangle$ and $\langle c+a \rangle$ slip systems. Twins are known to compensate this deficiency in Al lean alloys (0-2 wt.% Al) because of the decrease in the stacking fault energy. At the first instance, this appears to be consistent with the present results in Ti-407 alloy with low Al (0.85 wt %) where twins have been observed in several grains. However, it is intriguing to note that these twins are formed in soft grains where deformation is expected to occur primarily by a -type slip and not by twinning. It is possible that the higher strain rate imposed by the split-Hopkinson-bar test is responsible for the formation of the twins. Increased propensity for twinning has been known to occur in Ti alloys, especially with low Al content and higher strain and strain rate [7,21–23]. Alternatively, the twins could have developed from a slip transfer mechanism, previously shown to occur in CP-Ti where the prismatic slip in one grain has been shown to induce $\{10\bar{1}2\}$ deformation twinning in neighboring grain despite having a very low global Schmid factor [24]. The alignment of the twins across neighboring grains seen in Figure 3 is conceivably due to strong rolling texture present in this alloy and as a result twins propagate across similarly oriented grains with minimal interruption. The small offsets between the twins in neighboring grains is due to small misorientation between them.

While the origins of these deformation twins are still under investigation, it is equally intriguing to see a high density of $\langle c+a \rangle$ dislocations within the equiaxed grain where the deformation twin was also present. Pyramidal slip involving $\langle c+a \rangle$ dislocations have been shown to occur in Ti-Al crystals, but only in hard grains deformed along c -[0001] direction. Conversely, single crystal studies by Williams [6] and Cass [18] have shown that $\langle c+a \rangle$ dislocations are entirely absent in soft grains oriented for a -prism slip. In this context, soft grains that are mainly oriented for prism slip in this study that show the presence of $\langle c+a \rangle$ dislocations is rather surprising. $\langle c+a \rangle$ slip was also widespread within the alpha laths in the transformed portion of the duplex microstructure. It is noted that the exact orientation of these laths with respect to the applied stress axis is not determined at the present time and the effect of grain orientation cannot be ruled out. Nonetheless, the relative increase in the density of $\langle c+a \rangle$ dislocation in the alpha laths of the transformed region compared with the equiaxed alpha grains, despite the fact the slip length is much shorter in the former compared with the latter, points to possibility of aspects other than Schmid factor and CRSS values influencing the slip. It is speculated that the alpha/beta interface could be a potential source for $\langle c+a \rangle$ dislocations since these interfaces already contain a and c -type misfit dislocations. Efforts are currently underway to characterize the nature of these interfaces.

An important aspect of the deformation behavior of higher-Al content crystals, with particular emphasis on titanium alloys containing about 6 wt.% Al, is the occurrence of catastrophic failure at high strain rates [6,21] . This type of failure has not been observed in any split-Hopkinson bar tested samples of Ti-407 in this study, lending to the increased performance of this alloy at high strain rates compared with Ti-64. Based on the results from this study, the absence of catastrophic failure due to unstable shear in Ti-407 is most likely an effect of the increased deformation along the c -axis.

5. Conclusion

In this work, the deformation mechanisms in the alpha/beta titanium alloy TIMETAL® 407 were studied through split-Hopkinson bar testing. All samples in this study have been deformed in tension to failure. In this preliminary study the following conclusions can be made: Ti-407 deforms significantly by $\langle c+a \rangle$ slip despite the strong rolling texture present in the sample, the low Al-content in the alloy, and the higher CRSS for this slip system. It has also been shown that Ti-407 deforms by the $\{10\bar{1}2\}$ twinning mode. This is an expected deformation mode because the low Al content should reduce the stacking fault energy and promote twinning. The high degrees of strain obtained in this alloy is most likely due to the increased deformation along the c -axis.

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7. References

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