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In situ transmission electron microscopy deformation of the titanium alloy Ti–6Al–4V: Interface behaviour

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Abstract

In situ transmission electron microscopy (TEM) experiments have been carried out on the Ti–6Al–4V alloy. This work focuses on the dislocation motion close to the α/β interfaces. The interfaces influence the alloy strengthening by inhibiting the dislocation propagation but also by favouring the plastic deformation transmission. Results of deformation micromechanism dynamic are reported for the first time in this alloy at room temperature and compared to post-mortem TEM results.

Keywords: In situ TEM; α/β titanium alloy; Ti–6Al–4V; Dislocations; Interface

1. Introduction

Titanium alloys are used in a wide variety of applications, particularly in aeronautics because of their good mechanical properties and their low density. The most used of them is the Ti–6Al–4V alloy which is a two-phase alloy with a hcp α phase and a bcc β phase. Its microstructure depends on the thermomechanical treatments. The microstructure of the investigated alloy consists in primary nodules (α_P) and lamellar colonies (α_S/β) (Fig. 1a). Our observations indicate that the nodules volume fraction is close to the lamellar colonies one in agreement with recent results by Bridier et al. [1]. The details of a lamellar colony are illustrated in Fig. 1b: generally α_S laths are regularly separated by thin β laths (e.g. A and B in Fig. 1b) except where the β phase is discontinuous (e.g. C in Fig. 1b). Despite a very low volume fraction of the β phase, the dislocation movement is highly impeded by it. Then, the presence of this phase noticeably influences the mechanical properties.

In situ transmission electron microscopy (TEM) tensile tests were carried out on this alloy at room temperature. The dislocation dynamic and deformation micromechanisms under stress and temperature can only be studied by this technique. In the past it has only been performed on this alloy in cryogenic conditions (20 K) [2]. The role of the lamellar colony interfaces in the deformation is studied here and the present results are compared with post-mortem results found in the literature.

For the studied alloy, prismatic and basal glides of dislocations with a Burgers vector $\vec{\mathbf{b}}$ of type $a/3\langle 1\,1\,\bar{2}\,0\rangle$ (**a**-type) are the main deformation modes at room temperature [1], whereas the preferential deformation mode for α-titanium in this range of temperature is prismatic glide [3–5].

In titanium alloys, α–β interfaces in lamellar colonies have a near-Burgers orientation relationship (OR). The Burgers OR (1934) is:

$$(1\,0\,1)_\beta//(0\,0\,0\,1)_\alpha \text{ and } [1\,\bar{1}\,\bar{1}]_\beta//[2\,\bar{1}\,\bar{1}\,0]_\alpha.$$

Experimentally for a fully lamellar Ti-6242 alloy, a 0.7° misorientation is observed between one **a**-type direction and one bcc $\langle 1\,1\,1\rangle$ direction and a misorientation of 11.1° between another **a**-type direction and another bcc $\langle 1\,1\,1\rangle$ direction. The third **a**-type direction is not closely aligned with a $\langle 1\,1\,1\rangle$ direction but is close to a $\langle 1\,0\,0\rangle$ direction, with a misorientation of 5.96° [6]. The three **a**-type Burgers vectors of the hcp phase are then non-equivalent. Deformation transmission between the α and β phases will then depend on the **a**-type Burgers vector and the slip plane of the dislocations.

2. Experimental results

In situ tensile tests have been performed at room temperature with a Gatan straining holder. TEM observations were made

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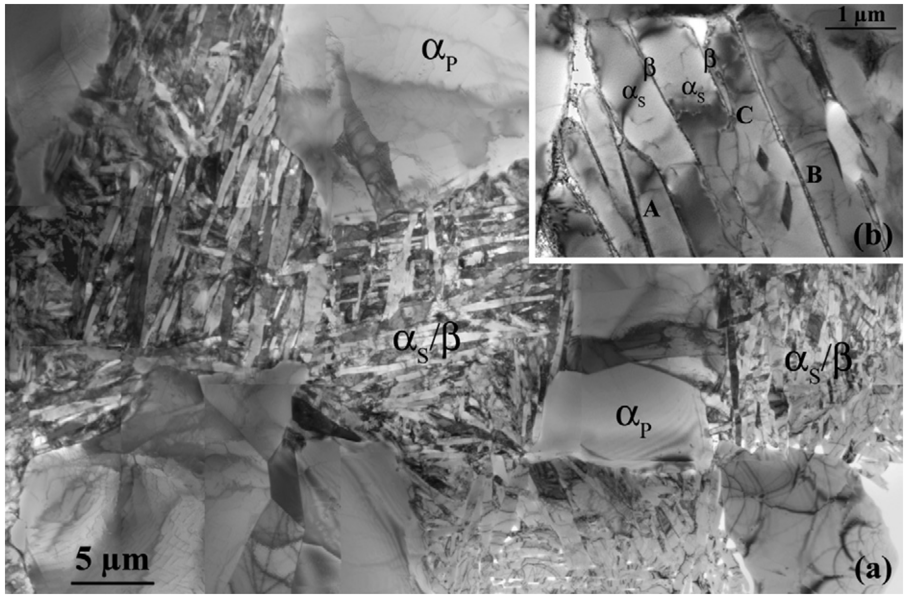

Fig. 1. TEM image of the general alloy microstructure (a) and detailed view of a lamellar colony (b).

with a JEOL 2010 equipped with a SIS CCD camera. Details of the TEM in situ deformation technique have been specified in previous studies [7].

The present work deals with two sequences recorded during in situ tests illustrating the interfaces–dislocations interaction.

Emission of dislocations from the α/β interfaces occurs very often, especially in the early stages of deformation. Fig. 2 shows an example of the emission of dislocations at a α_P/β interface. Two dislocations (noted 1 and 2 in Fig. 2a) are emitted from the interface by the local stress. On the next image (Fig. 2b) these two dislocations have slightly slipped and new dislocations are emitted from the interface. These dislocations have an **a**-type Burgers vector (determined by the $\mathbf{g}\cdot\mathbf{b}=0$ invisibility criterion) and glide in prismatic planes. The projection of the Burgers vector on the image shows that dislocation movement is controlled by screw segments, which are straight. At the end of the test (Fig. 2c) many dislocations have been emitted and are stopped in screw orientation. Stereographic projections of the α_P and β phases indicate that they have not a near-Burgers OR.

In another in situ sequence (Fig. 3) dislocations emitted at the interface α_S/β move in a lamellar colony. The study focuses on two α_S laths noted α_{s1} and α_{s2} separated by a thin β lath. A straight dislocation (arrowed in Fig. 3) with a screw character moves from right to left in the basal plane. Its Burgers vector has been determined as $\mathbf{b} = a/3[1\,1\,\bar{2}\,0]$ in the α structure. The two segments in the laths α_{s1} and α_{s2} slip generally well aligned, despite the presence of the β lath, while sometimes one segment is pinned and the other one continues to propagate (Fig. 3a). Their average velocity has been measured here to be $8\,\mathrm{nm\,s^{-1}}$. This dislocation behaviour has been often observed.

Simultaneous shearing of the three laths α_{s1}, α_{s2} and β is possible when it exists an OR between the two phases α_S and β. Then the Burgers vector $a/3[1\,1\,\bar{2}\,0]$ has a corresponding Burgers vector in the β phase equal or close to $\mathbf{b}'=a'/2[1\,1\,1]$, as suggested by Suri et al. [8] and Savage et al. [9]. The slip plane is common to the two phases (verified from the stereographic projections) and the angle between the two Burgers vectors **b** and **b**′ has been measured to be about 11°.

3. Discussion

The microstructure is constituted by α_P nodules and α_S/β lamellar colonies which are deformed together. Mechanisms observed at the interfaces allow the transmission of the defor-

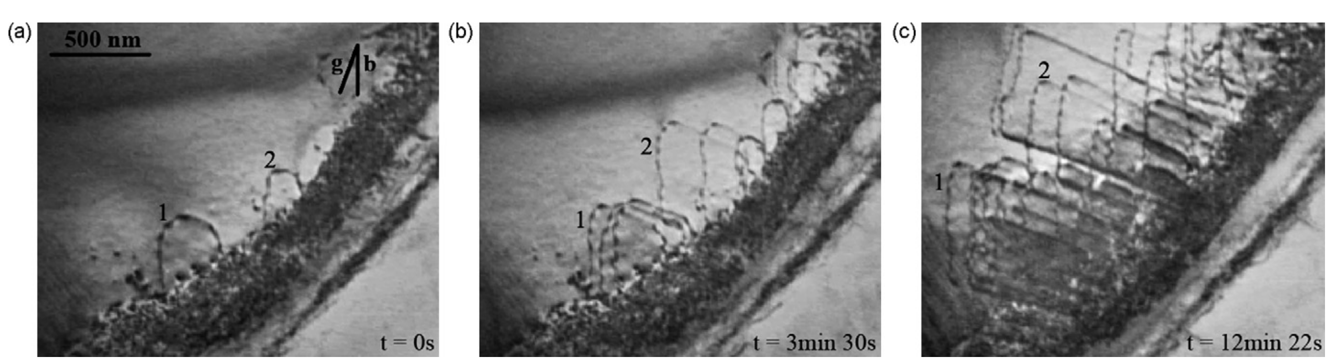

Fig. 2. In situ TEM observation of the dislocation emission from a β/α_P interface.

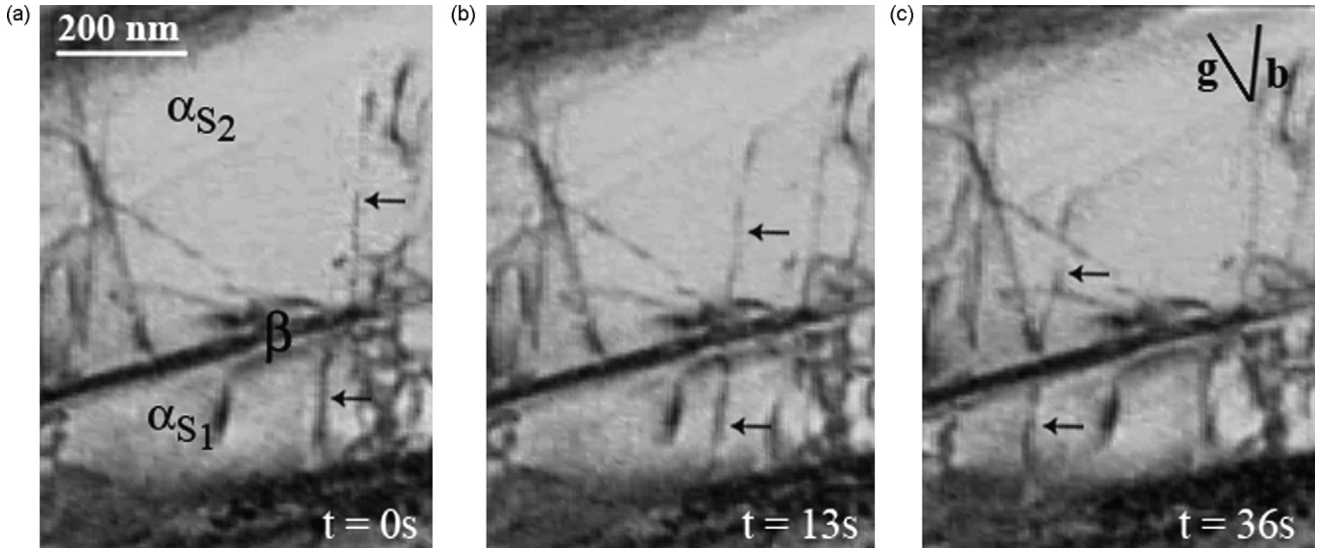

Fig. 3. In situ sequence showing simultaneous slip of dislocations in two α_S laths and one β lath.

mation and the compatibility of the deformation between the different phases. However, every interface type (β/α_S, β/α_P and α_S/α_P) has a different behaviour:

- In lamellar colonies, the OR allows the simultaneous deformation of α_S and β phases as presented here.
- Between colonies and α_P nodules, no special OR is observed and the situation is different every time. Therefore, dislocation emission is very often observed at β/α_P interfaces.

In the two sequences presented in this work, edge segments are moving quickly and the motion of screw segments controls the slip of dislocations in the α phase. This result is well known for titanium and its alloys: the three-dimensional core structure of screw dislocations implies that dislocations move by a locking–unlocking mechanism studied by many authors and with frequent cross slips [5,10–12]. Our results properly differ from those of Ambard et al. [2] on this point: at 20 K, they observed also basal slip of dislocations in lamellar colonies of Ti–6Al–4V but slip is rigorously planar in opposition to our observations at 300 K. This difference is certainly due to the thermal activation of the dislocation cross slip.

In the situation in Fig. 3, three laths α_{s1}, α_{s2} and β are deformed simultaneously. At the transmission of a dislocation through the α_S/β interface, a residual dislocation $\Delta\mathbf{b}$ is created. The reaction $\mathbf{b} \to \mathbf{b}' + \Delta\mathbf{b}$ occurs at the interface α_{s1}/β (with $\mathbf{b}$ and $\mathbf{b}'$, respectively, the Burgers vectors in the α and the β phase) and the inverse reaction $\mathbf{b}' \to \mathbf{b} - \Delta\mathbf{b}$ occurs at the interface β/α_{s2}. These transmission reactions have been already proposed by Savage et al. for a fully lamellar Ti-6242 alloy [9], where the OR in lamellar colonies is the same as in Ti–6Al–4V alloy. Our experiments show that the residual dislocations have a small modulus length (about $0.2\|\mathbf{b}\|$) and make possible the coordinated motion of dislocations in several different phases. Geometric considerations show that residual dislocations have a character close to the screw orientation and are then mobile in the interface plane. Therefore, they are able to combine and it has been shown that another **a**-type dislocation can be formed from six residual dislocations as suggested by these same authors. After the in situ test, some dislocations with an **a**-type Burgers vector different from $a/3[1\,1\,\bar{2}\,0]$ have been observed close to the interface. They could be the result of such interaction; but due to the high density of $a/3[1\,1\,\bar{2}\,0]$ dislocations, it was not possible to determine exactly their Burgers vector.

As dislocations move approximately with the same velocities in the two laths α_{s1} and α_{s2}, while being rectilinear, our experiments show that the presence of the β phase does not inhibit their movement and the deformation is essentially controlled by motion of screw dislocations in the α phase. Among the three possible **a**-type Burgers vectors, the one observed produces the larger residual dislocation. Energetically, it is the more unfavourable slip system but it has the highest resolved shear stress for the orientation of our test, this is why it is observed here.

Finally, effective stresses close to the α_S/β interfaces can be different from the resolved shear stress. We have evaluated the local stress close to the interface [13]: it increases from 80 to 220 MPa, respectively, from the α_S lath centre to a distance at 100 nm from the interface. This difference is probably due to the compatibility stresses between the α and β phases as reported by Ankem et al. [14,15]. It could be an explanation for our observations of many dislocations movements and emissions close to the α/β interfaces.

4. Conclusion

Previous results from post-mortem observations in Ti–6Al–4V have been confirmed here by TEM in situ deformation tests:

- the α phase deformation is controlled by the screw segment movements of **a**-type dislocations;
- some slip systems allow the deformation transmission between the α and β phases in lamellar colonies;

- in the case of transmission, residual dislocations are created at the α/β interfaces.

In addition, this study shows clearly that dislocations are emitted from the α/β interfaces at the early stage of the deformation, that coordinated movement of dislocations in different phase laths is possible and that β phase does not inhibit the dislocation movement in lamellar colonies.

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