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In situ SEM study of tensile deformation of a near-β titanium alloy

Saifei Zhang[a,b,*], Weidong Zeng[b,**], Qinyang Zhao[c], Liling Ge[a], Min Zhang[a]

[a] School of Materials Science and Engineering, Xi'an University of Technology, Xi'an 710048, China
[b] State Key Laboratory of Solidification Processing, Northwestern Polytechnical University, Xi'an 710072, China
[c] Waikato Centre for Advanced Materials, School of Engineering, The University of Waikato, Hamilton 3204, New Zealand

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ABSTRACT

In this study, the deformation behavior of a near-β Ti-17 titanium alloy under uniaxial tension was investigated using electron backscatter diffraction (EBSD) and tensile tests with *in situ* scanning electron microscope (SEM) observation. It is found slip mode is the main deformation mechanism in primary α grains of Ti-17 during the tensile test. Slip systems activated were identified by performing calculations on EBSD orientation data. The results show that all the three slip systems with **a**-type Burgers vector, *i.e.* basal, prism and 1st-order pyramidal slip, could be activated in the primary α grains of Ti-17, but no **a** + **c**-type slip (2nd-order pyramidal slip) activated is observed. Analysis reveals that basal and prism slips are the dominating slip mode, while 1st-order pyramidal slip acts as a subsidiary or deviated slipping mode. For the equivalent slip systems, Schmid factor dominates the slip behavior, while for the non-equivalent slip systems, critical resolved shear stress (CRSS) must be taken into account. It is proved that CRSS for **a** + **c**-type slip is much larger than that for **a**-type slips (basal and prism slips) in Ti-17 alloy (at least 2.5 times that for basal slip). **a**-type slip remains the easiest slip to be activated even in the condition favoring **a** + **c**-type slip and suppressing **a**-type slip (an angle of ~10° between the tensile axis and ***c***-axis).

1. Introduction

Titanium and titanium alloys are excellent candidates for applications in aerospace, marine engineering and biomedical field due to their high strength to weight ratio, excellent corrosion resistance and biological compatibility [1–3]. In most of the applications, titanium alloys are used as important load-bearing parts, for which deformation and fracture of the material is the main concern. It has been reported in titanium alloys that mechanical properties of these alloys are greatly influenced by their metallurgical microstructure [2]. In order to get the complete overview of the mechanical properties of titanium alloys, it is necessary to understand the deformation nature at microstructural level when a mechanical loading is applied.

Considerable efforts have been devoted to the study the deformation behavior of Ti alloys. Most of the studies concerned the workability of the material at high temperatures [4,5], or the experimental data reported concerned mainly the macro elastic/plastic responses at different strain rates and temperatures [6,7]. Many studies have been devoted to the understanding of the specific plastic deformation mechanisms at room temperature in pure titanium [8–11], while works on understanding the deformation behavior of Ti alloys from perspective of microstructural aspects is rather limited [12]. By reviewing the literature, it is found that the deformation of hexagonal titanium, unlike the cubic metals, can be more complex. Plastic deformation modes of α-Ti at low temperatures include both conventional slip by dislocations and twinning deformation mode [10–14]. Twinning mode is important for the deformation of pure-Ti and some α titanium alloys with coarse grain structure [13]. But in α + β titanium alloy, twining mode is nearly completely suppressed because of fine microstructure and high solute contents [14]. So the slipping behavior is the core issue for its deformation, ductility and fracture. For the slip mode in α-Ti, the main slip directions are the three close packed directions $< 11\bar{2}0 >$ (**a**-type Burgers vector). The slip planes containing $< 11\bar{2}0 >$ directions are the basal plane (0001), the three prism planes $\{10\bar{1}0\}$ and the six 1st-order pyramidal planes $\{10\bar{1}1\}$. So there are a total of 12 **a**-type slip systems, which are the three basal slip systems $(0001) < 11\bar{2}0 >$, three prism slip systems $\{10\bar{1}0\} < 11\bar{2}0 >$ and six 1st-order pyramidal slip systems $\{10\bar{1}1\} < 11\bar{2}0 >$. Additionally, there is a **c** + **a**-type slip mode that has been observed by TEM method in some titanium alloys with slip system of $\{11\bar{2}2\} < 11\bar{2}3 >$ [15], which is called 2nd-order pyramidal slip systems. It should be noted that the critical resolved shear stress (CRSS) for **c** + **a**-type slip is significantly larger than **a**-type slip, so it is possible

* Corresponding author at: 127 West Youyi Road, Xi'an Shaanxi 710072, P.R.China.
** Corresponding author.
E-mail addresses: zhsaifei@mail.nwpu.edu.cn (S. Zhang), zengwd@nwpu.edu.cn (W. Zeng).

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be activated only in the condition that the applied stress is parallel to the c-axis [14]. The fundamental research provides useful guidance for the present work.

On the other hand, tremendous progress has been made in material analysis and testing technology in recent years. Electron backscatter diffraction (EBSD) is now being increasingly utilized for the analysis of crystal orientations at micro-level. Mechanical test can be carried out with *in situ* SEM observation, making it possible to record any microstructure changes during loading. The combination of these techniques has been applied in multiple materials [16,17], which provide references for the experimental design in the present work.

By using EBSD techniques and *in situ* SEM observation of tensile deformation, the present work aims to investigate the tensile deformation behavior of a near-β titanium alloy at microstructural level. This article focuses on the slip mode analysis using a method developed for determining the activated slip systems.

2. Material and experimental procedures

2.1. Material

The material used in the present work with the nominal composition of Ti-5Al-2Sn-2Zr-4Mo-4Cr (wt%) was cut from an α + β processed disk forging. The material is forged at α + β phase region followed by subtransus solution treatment and ageing (STA) heat treatments. The microstructure is shown in Fig.s 1 and 2. It is the so-called bi-model structure, which consists of equiaxed primary α (α_p) grains with diameter of 5–8 μm dispersed in the transformed β matrix. The volume fraction of primary α phase is about 35%, which is largely determined by forging temperature. The transformed β matrix is composed of fine acicular α at different scales formed during cooling and heat treatments.

2.2. EBSD measurements on the in situ tensile sample

Two samples were prepared for tensile test by electrical-machining with overall dimensions of 35 mm × 15 mm × 1 mm and gauge width of 3 mm. They were electro-polished to remove the residual stress and improve surface quality. Area of Interest (AOI) was then marked by micro indentation at the center of the gauge area for the convenience of subsequent *in situ* observation. EBSD measurements were conducted on AOI to acquire the crystallographic information for the samples. HKL-Channel 5 software is used to do crystal orientation analysis.

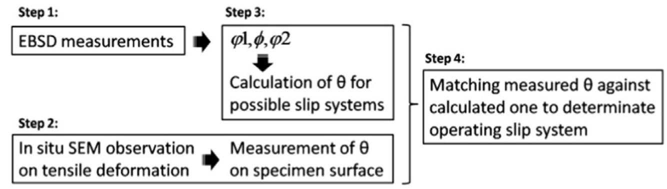

Fig. 2. The steps for determining the slip system activated in HCP α-Ti.

2.3. Tensile test with in situ SEM observation

Uniaxial tensile experiment was conducted in the vacuum chamber of Zeiss Supra 55 FEG scanning electron microscope (SEM) using a Gatan servo-hydraulic loading system. The test system allows interrupted stops to observe the possible microstructural changes. Before the test, chemical etching was performed in order to acquire high quality microstructural images. In order to ensure the stability of the system, loading speed of 0.1 mm/min is used in this study.

3. The methodology for determining the slip systems activated in HCP α-Ti

In the present study, the method for determining slip systems activated in α-Ti includes 4 steps, as shown in (Fig. 2): First, EBSD technique is employed to obtain the crystallographic orientation data in the area of interest (AOI), i. e. the three Euler angles. Second, *in situ* SEM observation under uniaxial tension is conducted, and the angles between the slip lines and loading axis are measured. Third, the orientation data obtained in the first step is used to calculate the angles between the slip lines and loading axis for possible slip systems in α-Ti. Fourth, match the measured angle with the calculated one to determinate the operating slip system.

In the 4 steps, step 3 is the key step. In this study, 18 slip systems in α-Ti are considered (Fig. 3), including 3 basal slips, 3 prism slips, 6 first-order pyramidal slips and 6 s-order pyramidal slips. When a crystal is under stress, not every slip systems are operated. The selection of 'possible slip systems' in step 3 is based on the classical Schmid law [18], which postulates that a critical value (CRSS) of the resolved shear stress (τ) must be reached, if a given slip system is activated in a deformed crystal. Schmid law has proved to be a very useful tool and scarcely questionable in most cases [19,20], even though there are a few reports of breakdown of Schmid law in very specific situations [21].

The stress state analysis when a slip system is activated is shown in Fig. 4. In the figure, λ is the angle between stress axis and slip direction; $\varphi$is the angle between stress axis and the normal direction of slip plane. The component shear stress of σ in the slip direction, *i.e.* τ, is determined as follows:

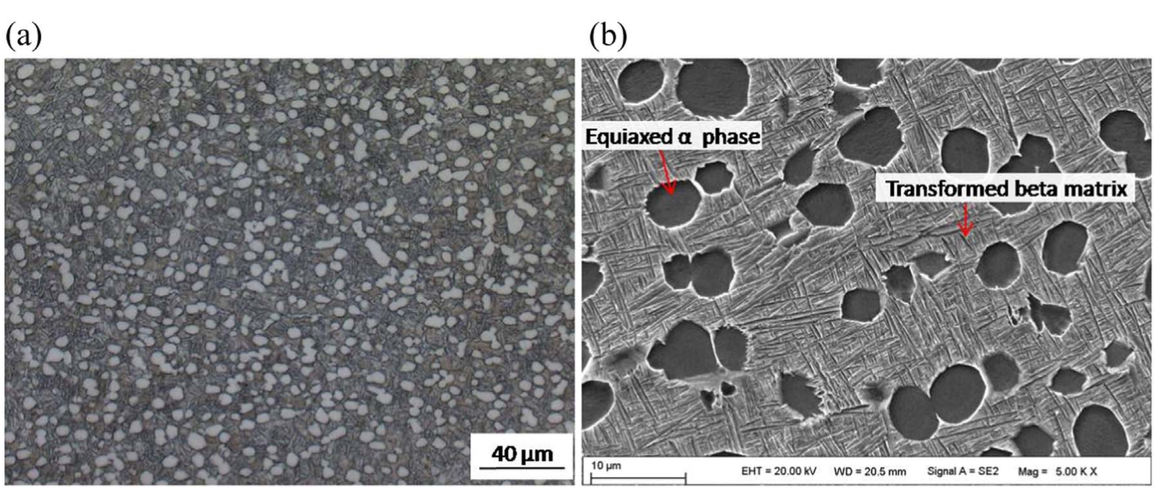

Fig. 1. Microstructure of Ti-17 alloy used in the present study by optical microscopy (a) and SEM microscopy (b).