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Deformation induced face-centered cubic titanium and its twinning behavior in Ti–6Al–4V

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ABSTRACT

The transformation of crystal structure and twinning behavior for Ti–6Al–4V subjected to severe plastic deformation were studied. The face-centered cubic titanium in Ti–6Al–4V and the deformation twins in face-centered cubic titanium of Ti–6Al–4V were found via transmission electron microscopy and high-resolution transmission electron microscopy. And the deformation twins existed in the grain size of a few nanometers for equiaxed nanograins to about 150 nm in width for elongated ultrafine grains. In addition, the twinning mechanism was the partial dislocation emission from the grain boundary during the deformation of nanocrystalline face-centered cubic titanium.

Nanocrystalline (NC) materials are well known for their high hardness and strength, good tribological properties, enhanced electrical resistivity, higher thermal expansion coefficient, higher heat capacity and so on [1], but frequently suffer significant losses in ductility [2]. Twins in NC materials have been reported to increase both the strength and the ductility by accumulating dislocations at the twin boundaries [3,4]. Titanium (Ti) and its alloys have attracted great attention with regard to their comprehensive applications [5,6]. It is very promising and attractive to induce the twins in NC of Ti and its alloys for further improving their corrosion resistance, wear resistance and fatigue performance.

It is well known that twinning is a common and important phenomenon during the plastic deformation of metals and alloys. Except for the treatment temperature and the type of alloy [7], the grain size plays a key role in dominating twinning behavior of metals and alloys [8]. For hexagonal close-packed (hcp) Ti, a decrease in grain size ranging from the CG scale to the ultrafine grain (UFG) scale monotonously reduces the propensity of twinning [9], including a transition from deformation twinning to ordinary slip-dominated processes, and the twins have never been observed in the NC hcp-Ti [10]. However, the traditional face-centered cubic (fcc) metals and alloys are more difficult to deform by twinning in a smaller coarse-grain (CG) [11], but twinning becomes easier as the grain size decreases to a few decade nanometer [12]. Meanwhile, the fcc-Ti is found to be operative in Ti and its alloys [13–15]. Nevertheless, the previous works present that the deformation twins have not been observed in any case of fcc-Ti. Whether can the twinning occur in fcc-Ti, and what are the mechanism and the scale in the twinning fcc-Ti? In this letter, we will demonstrate the experimental investigation of twinning behaviors under severe plastic deformation (SPD) for Ti–6Al–4V (the most widely used titanium alloy).

In this work, Ti–6Al–4V consisting of predominantly high volume fraction of hcp α phase with a grain size of 9.37 μm and a complementary small amount of β phase distributing in α grain boundaries was adopted. The chemical composition (wt.%) of as-received Ti–6Al–4V bar with a diameter of 40 mm is composed of 6.41Al, 4.19V, 0.02Fe, 0.006C, 0.001N, 0.16O, 0.002H and Bal. Ti. The annealing treatment of Ti–6Al–4V at 873 K for 1 h was followed to cool in air to room temperature, and the Ti–6Al–4V specimens with a dimension of 70 mm × 19 mm × 4 mm were manufactured from the as-annealed Ti–6Al–4V and then Ti–6Al–4V specimens were polished with silicon carbide paper to grade 600. Subsequently, Ti–6Al–4V specimens were subjected to high energy shot peening (HESP) at room temperature, in which HESP was performed on MP6000PT air blast machine at an air pressure of 0.25 MPa and processing duration of 60 min. In addition, the standard cast steel shots ASH230 with a diameter of 0.6 mm were stored and the mass flow rate of about 10 kg/min was adopted. Transmission electron microscopy (TEM) and high-resolution TEM (HRTEM) observations were conducted on a Tecnai F30 G2 field emission transmission electron microscope at a voltage of 300 kV. The nanograins and ultrafine grains in the treated surface layer with a thickness of about 50 μm for the bulk Ti–6Al–4V after HESP were checked, and the grain size decreased with the decreasing of depth below the topmost surface

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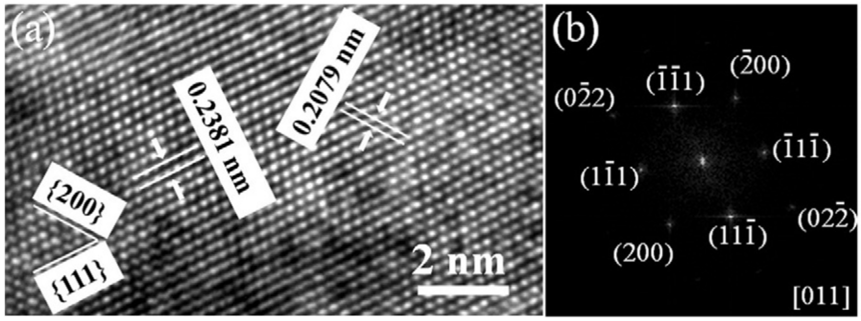

Fig. 1. (a) HRTEM image of the approximately distortionless region at about 50 μm below the topmost surface; (b) the FFT image corresponding to (a).

accompanied by an increase in strain and strain rate. Therefore, the HESP processed Ti–6Al–4V offered sufficient samples to investigate the transformation of hcp-Ti with different grain sizes and the deformation behaviors of fcc-Ti.

It is generally understood that phase transformation leads to a variation of the lattice parameters, and for hcp-Ti, the lattice parameters are $a = 0.2945$ nm and $c = 0.4648$ nm ($c/a = 1.578$). The fine structures of the approximately distortionless region at about 50 μm below the topmost surface for Ti–6Al–4V after HESP are observed via HRTEM as shown in Fig. 1(a). Fig. 1(b) reveals that the crystal structure in this region is in fact fcc structure as verified by using the fast-Fourier-filtered (FFT) image taking along [011] zone axis. The interplanar distances of {200} and $\{\bar{1}1\bar{1}\}$ planes are determined to be 0.2079 nm and 0.2381 nm respectively as seen from Fig. 1(a), and signifying that the lattice parameter of the fcc phase is calculated to be 0.4158 nm. The result is consistent with the previous results that the lattice parameter of fcc-Ti is mostly in the range of 0.41–0.43 nm [13–16]. The aforementioned results have strongly proved that fcc-Ti has been induced in the HESP process of Ti–6Al–4V due to SPD. It is widely accepted that the phase transformation is a stress-induced process and strongly influenced by the strain rate and grain size [13]. Meanwhile, it is commonly understood that the strain rate at the processed surface is extremely high during HESP [17], and therefore fcc-Ti can be activated in the hcp-Ti. In addition, the deformation induced transformation has been clearly investigated and is attributed to the gliding of Shockley partial dislocations with a Burgers vector of $\frac{a}{6}\langle 11\bar{2}0\rangle$ [14], where a is the lattice parameter of fcc-Ti.

Further observations suggest that the twinning occurs in fcc-Ti of Ti–6Al–4V. As seen from the bright-field (BF) and dark-field (DF) TEM images in Fig. 2(a) and (b), the grain has been obviously refined. Meanwhile, the high-density parallel twins are observed in the ultrafine plate-shaped grain about 150 nm in width as shown in Fig. 2(a) and (b), presumably due to the higher stress. The selected area electron diffraction (SAED) pattern shown in the inset of Fig. 2(a) confirms that the grain is with fcc structure due to the fact that the zone axis is [011] and the twinning plane is identified as $(\bar{1}1\bar{1})$ as confirmed by the schematic shown in Fig. 2(c).

It should be noteworthy that the grain shown in Fig. 2(a) is still in the UFG scale. The deformation twinning for traditional fcc metals, such as Al, Cu, and Ni difficultly occurs under such a grain size, but easily does that in a nanometer scale [11,12,18,19]. For this reason, it is necessary to determine whether the twinning is induced in the NC fcc-Ti, and how the twining occurs in the NC fcc-Ti.

As expected, we do have found the twins in NC fcc-Ti as shown in Fig. 3. Three twin boundaries (TBs) indicated by using TB_a, TB_b and TB_c are in a 25 nm × 40 nm NC as shown in Fig. 3(a), and the dash dotted lines mark the grain boundary (GB). The three TBs are approximately parallel arrangement. The corresponding FFT image shown in Fig. 3(b) clearly reveals splitting spots due to the presence of twins. Since mirror spots appear with respect to $(\bar{1}1\bar{1})$ plane, NC is also with fcc structure, as verified by FFT image shown in Fig. 3(b). The inverse fast Fourier-filtered (IFFT) image corresponding to the white frame in Fig. 3(a) is shown in Fig. 3(c). There is only one twin boundary TB_a as indicated in the lower right part of the image that the twin is divided into domains I and II. As seen from Fig. 3(c), some microtwins and stacking faults (SFs) are also observed in the upper left part of I as indicated by the white line, and the microtwins and SFs do not cross the whole NC and stop in the grain interior with Shockley partial dislocations. It is obvious that these microtwins are heterogeneously nucleated at GB and grow into the grain interior via partial dislocation emission from GB.

We have not only found the deformation induced twins in the bigger NC, but also did the twins existing in the extremely tiny NC fcc-Ti. As

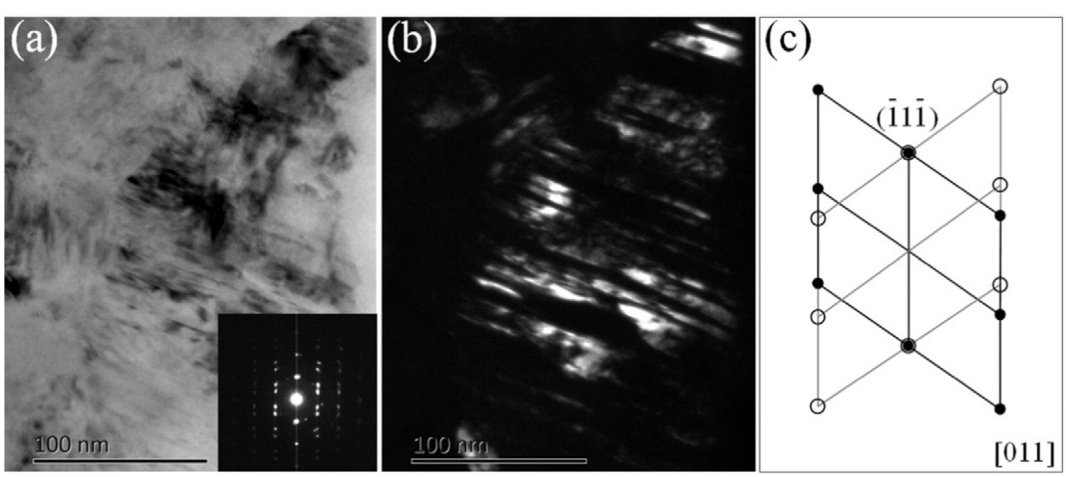

Fig. 2. The typical TEM images at about 50 μm below the topmost surface: (a) the BF image and the inset showing the corresponding SAED pattern; (b) the DF image; (c) schematic of the SAED pattern in (a).