

Article

In Situ Observation of Tensile Deformation of Ti-22Al-25Nb Alloy and Characterization of Deformation in α_2 Phase

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Abstract: The room temperature tensile deformation of Ti-22Al-25Nb alloy with an equiaxed α_2 phase microstructure and the activated slip system of α_2 particles were investigated by a combination of in situ tensile tests and electron backscatter diffraction experiments. The results demonstrate that only a few wide and long slip bands occur in the B2 matrix in the initial stage of yielding. With the tensile displacement increases, a large number of slip bands, including multiple- and cross-slip bands, appear in the B2 matrix and the distance between two adjacent slip bands decreases significantly. Meanwhile, the movement of the slip bands is hindered by the α_2 particles and the B2 grain boundaries, and the slip bands appear only in a small number of the α_2 particles. From the beginning of the tensile process to the final fracture, there are lots of α_2 particles without slip bands. The slip bands penetrate the needle-like lamellar O phase without changing the slip direction. Compared with the α_2 particles, the hindering effect of needle-like O phases on the motion of the slip bands is quite small. The microcracks nucleated at the α_2 /B2 phase boundaries or within the α_2 particles, and microcracks propagated along the α_2 /B2 phase boundaries or across the α_2 particles. The fracture surface shows the quasi-cleavage feature, which contains a large number of small and shallow dimples on planar facets. The analysis indicates that the plastic deformation of the alloy is mainly contributed by the B2 phase. For room temperature tensile deformation of α_2 phase, there are three types of slip systems that can be activated, including the prism $\langle a \rangle$ type slip, the basal $\langle a \rangle$ type slip and the pyramidal $\langle a+c \rangle$ type slip. The prism $\langle a \rangle$ type slip is most likely to be activated, followed by the basal $\langle a \rangle$ type slip and finally the pyramidal $\langle a+c \rangle$ type slip. In addition, the critical resolved shear stress (CRSS) for the pyramidal $\langle a+c \rangle$ type slip is the highest among the three types of slip systems. Therefore, the deformation in the α_2 phase is mainly contributed by the prism $\langle a \rangle$ type slip and the basal $\langle a \rangle$ type slip.

Keywords: Ti₂AlNb-based alloy; equiaxed α_2 particles; in situ tensile observation; slip system; deformation characteristics



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1. Introduction

Since the first generation of the Ti₂AlNb-based alloy, Ti-25Al-12.5Nb (at.%), was discovered by Banerjee [1]; Ti₂AlNb-based alloys have attracted a great deal of interest in the aerospace field as prospective elevated-temperature structural materials due to their high specific strength, good creep resistance and oxidation resistance [2–4]. As a second generation of Ti₂AlNb-based alloy, the Ti-22Al-25Nb alloy has an excellent balance of room temperature and elevated-temperature properties [5–7]. Therefore, a lot of studies on

this alloy have been conducted, such as microstructural evolution, phase diffusion and mechanical properties [8–10].

There are generally three phases in the Ti_2AlNb -based alloy, including the B2 phase as an ordered body-centered cubic structure phase, the α_2 phase as a hexagonal close-packed (base on Ti_3Al) structure phase and the O phase as an orthorhombic (based on Ti_2AlNb) structure phase [11]. The mechanical properties of Ti_2AlNb -based alloys are significantly influenced by the microstructures due to the diversity of the phase structure. Boehlert [12] suggested that Ti-25Al-24Nb alloys with single B2 phases presented a poor tensile property (ultimate tensile strength ≤ 672 MPa, elongation $\leq 0.6\%$), and those with fully O phases exhibited better mechanical property (ultimate tensile strength ≤ 704 MPa, elongation $\leq 1\%$) than those with fully B2 microstructure. The tensile deformation of Ti-22Al-25Nb with different sizes of lamellar O microstructures was investigated by Zeng et al. [13–15]. The ultimate tensile strength and elongation ranged from 873–1065 MPa and 6.5–14.8%, respectively. Wang et al. [16] studied the tensile behavior of a Ti-22Al-25Nb (at.%) orthorhombic alloy with an equiaxed O/ α_2 + B2 microstructure and found that the alloys with this microstructure had favorable tensile properties (ultimate tensile strength ≤ 1180 MPa, elongation $\leq 8\%$). In addition, the deformation mechanisms of the different constituent phases have a serious influence on the tensile properties of the alloy. There are a large number of studies focused on the deformation mechanisms of the B2 phase and O phase. The B2 phase contains three different slip systems, which are $\{110\}\langle 111 \rangle$, $\{112\}\langle 111 \rangle$ and $\{123\}\langle 111 \rangle$, respectively [17]. The critical resolved shear stress of $\{123\}\langle 111 \rangle$ is higher than that of the other two [18]. Kamat et al. [19] proposed that the B2 phase contributes to the blunting of microcracks and improves the fracture toughness of the alloy. Banerjee et al. [20] investigated the deformation of the O phase in Ti-Al-Nb alloy and found that $[100]$, $[110]$, $[102]$ and $[114]$ are the four slip directions that can be activated at room temperature and elevated temperature. Poplille et al. [21] found that only two types of dislocations were activated in the O phase at room temperature. Zhang et al. [22] found that the O phase and the B2 phase have good slip compatibility and basal $\langle a \rangle$ type slip, prism $\langle a \rangle$ type slip and pyramidal $\langle c+a \rangle$ type slip of the O phase can be activated at room temperature. Studies on activated slip systems for the α_2 phase have mainly focused on Ti_3Al alloys. There are three types of dislocation for α_2 - Ti_3Al [23]: (I) the basal $\langle a \rangle$ type with a Burgers vector of $\langle 11\bar{2}0 \rangle$ moving on basal (0001), prism $\{10\bar{1}0\}$ or pyramidal $\{20\bar{2}1\}$ planes, (II) $\langle c \rangle$ type with a Burgers vector of $[0001]$ on second-order $\{11\bar{2}0\}$ prism planes and (III) $\langle c+a \rangle$ type with a Burgers vector of $\langle 11\bar{2}6 \rangle$ on $\{11\bar{2}1\}$ or $\{20\bar{2}1\}$ pyramidal planes. The absence of c-component deformation results in the generation of local stresses resulting in brittle cleavage and poor ductility. Djanarthany et al. [24] summarized the mechanical properties and deformation mechanisms of monolithic titanium aluminides based on Ti_3Al . They proposed that the absence of $\langle c \rangle$ slip and the low mobility of $\langle c+a \rangle$ dislocations contribute to stress concentrations at grain boundaries and twinning deformation is not possible because it leads to a disorder incompatible with the α_2 - Ti_3Al structure. Court et al. [25] investigated the influence of alloying additions on the mechanisms of plastic deformation of Ti_3Al alloy containing 4 at.% Nb. They found that the $\langle 11\bar{2}0 \rangle$ and $\langle 11\bar{2}6 \rangle$ dislocations become more active with the addition of niobium elements and the pyramidal $\langle c+a \rangle$ type slip system is activated at room temperature. Banerjee et al. [26] proposed that the increase in the Nb content of the α_2 phase leads to increasing basal glide of $\langle a \rangle$ dislocations and no pyramidal $\langle a \rangle$ slip was activated during room temperature deformation. Legros et al. [27] studied the deformation process in a Ti-23.7Al-9.4Nb alloy using in situ experiments performed at room temperature. They suggested that the main difference between the α_2 phase in Ti-23.7Al-9.4Nb alloy and the single-phase Ti_3Al alloy is the higher homogeneity of the basal glide in the α_2 phase. As mentioned above, the slip mechanism of the α_2 phase changes considerably with the addition of β -stabilizing elements. However, there are relatively few studies on the α_2 phase slip deformation mechanism of the Ti-22Al-25Nb alloy. Therefore, it is particularly necessary to study the activated slip systems of the α_2 phase due to higher Nb content in the Ti-22Al-25Nb alloy.

This paper reports the research on the room temperature deformation of the α_2 phase in Ti-22Al-25Nb (at.%) alloy. The aim of this work is to investigate the deformation characteristics of the α_2 phase using in situ tensile tests and electron backscatter diffraction experiments. This information will be beneficial to designing high-performance Ti-22Al-25Nb alloy components in practical production.

2. Materials and Methods

2.1. Material

The as-received Ti₂AlNb-based alloy bar was supplied by the High Temperature Materials Research Institute of Central Iron and Steel Research Institute (Beijing, China). Its nominal and chemical composition are shown in Table 1. The chemical composition was determined by inductively coupled plasma atomic emission spectrometry (ICP-AES). The bar was isothermally forged at (O + B2) phase region, and the microstructure after forging is shown in Figure 1. As can be observed in Figure 1b, the microstructure of the forged alloy is composed of the equiaxed α_2 /O particles, coarse lamellar O phase and B2 matrix. The equiaxed α_2 /O particles consist of α_2 particles and rim O, which are randomly and homogeneously distributed in the B2 matrix. In order to better investigate the deformation behavior of the α_2 phase in the Ti-22Al-25Nb alloy during the room temperature tensile process, the alloy was solution treated at 1020 °C for 30 min followed by water quenching. The microstructure of the as-received alloy was observed using optical microscopy (OM, OLYMPUS/PMG3, Olympus Corporation, Tokyo, Japan), and scanning electron microscope (SEM, ZEISS SUPRA55, Carl Zeiss, AG, Jena, Germany). The SEM images were analyzed using the Image-Pro Plus 6.0 software (Media Cybernetics, Rockville, MD, USA) to measure the volume fraction and size of the α_2 particles. The volume fraction of α_2 particles is approximately 33%. After statistical analysis, the diameter distribution of α_2 particles is in accordance with the normal distribution, and the average diameter of α_2 particles is 5.2 μ m. The 95% confidence interval of the average diameter is 5.05 μ m, 5.38 μ m.

Table 1. Chemical composition of the as-received alloy (at.%).

Nominal	Analysis					
	Ti	Al	Nb	O (ppm)	N (ppm)	H (ppm)
Ti-22Al-25Nb	Bal.	22.3	25.7	430	52	9

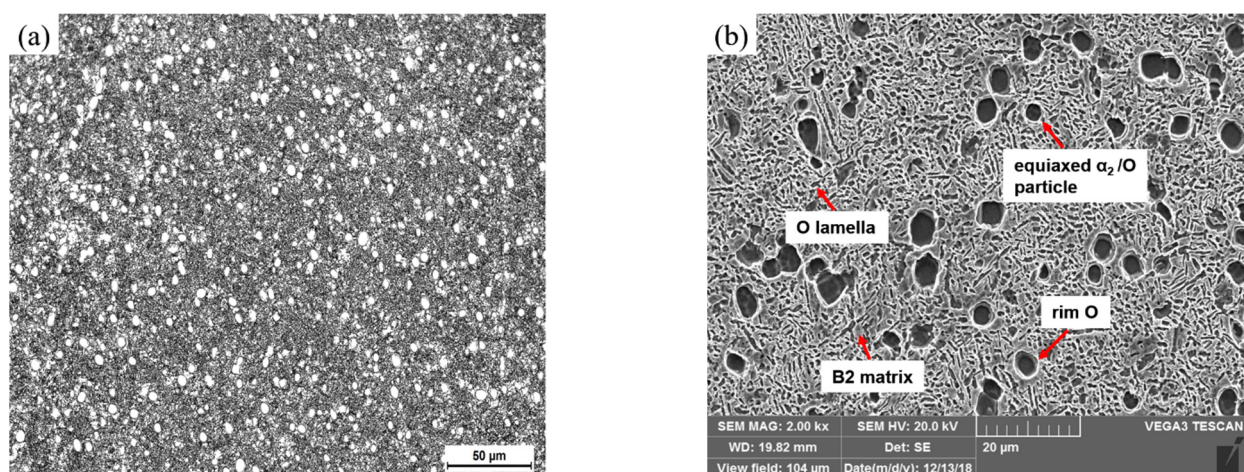


Figure 1. Microstructure of the alloy after forging: (a) OM; (b) SEM.

2.2. In Situ Tensile and EBSD Experiment

The in situ tensile test with electron backscatter diffraction (EBSD) was conducted in the present work to examine the deformation and activated slip systems of α_2 phase in Ti-22Al-25Nb at room temperature. The specimens for the in situ tensile test and

EBS D experiment were cut from the solution-treated bar using a wire electrical discharge machine (DK7732HB, Posittec Equipment, Suzhou, China), coarsely ground on with 80# metallographic sandpaper, then finely ground with 1000# metallographic sandpaper and finally electrolytic polished with an electrolyte of 10% perchloric acid, 35% n-butyl alcohol and 55% methanol in volume fraction. The electrolytic polishing process lasted for 40 s at a voltage of 40 V and a current of under 1 A at 15 °C. The geometric dimensions of the specimen for the in situ tensile test are shown in Figure 2. The gauge dimensions of in situ tensile specimens are 1.58 mm in length and 1.8 mm in width. The following experiments consist of two steps. For the first step, an area of interest (AOI) near the center of the gauge region of the specimens was selected for EBS D experiments on ZEISS Gemini 500 field emission scanning electron microscope (Carl Zeiss AG, Jena, Germany). The specimens were positioned on a 70° sample stage. The step size, testing voltage and resolution rate of the EBS D analysis were 0.2 μm , 20 kV and 97.84% respectively. The EBS D data were analyzed using HKL-Channel 5 software (HKL, Oxford, UK). For the second step, the specimens used in the EBS D experiments were etched with a corrosive of 11% HF, 33% HNO_3 and 56% H_2O in volume fraction and then placed in the vacuum chamber of the ZEISS Gemini 500 field emission scanning electron microscope with a servo-hydraulic loading platform (Gatan, Pleasanton, CA, USA) for room temperature in situ uniaxial tensile SEM observation. The rate of tensile displacement is 0.1 mm/min. The loading platform will hold the load during each temporary pause in displacement for SEM observation. To ensure the reliability of the experiment, three specimens were used for the in situ uniaxial tensile test.

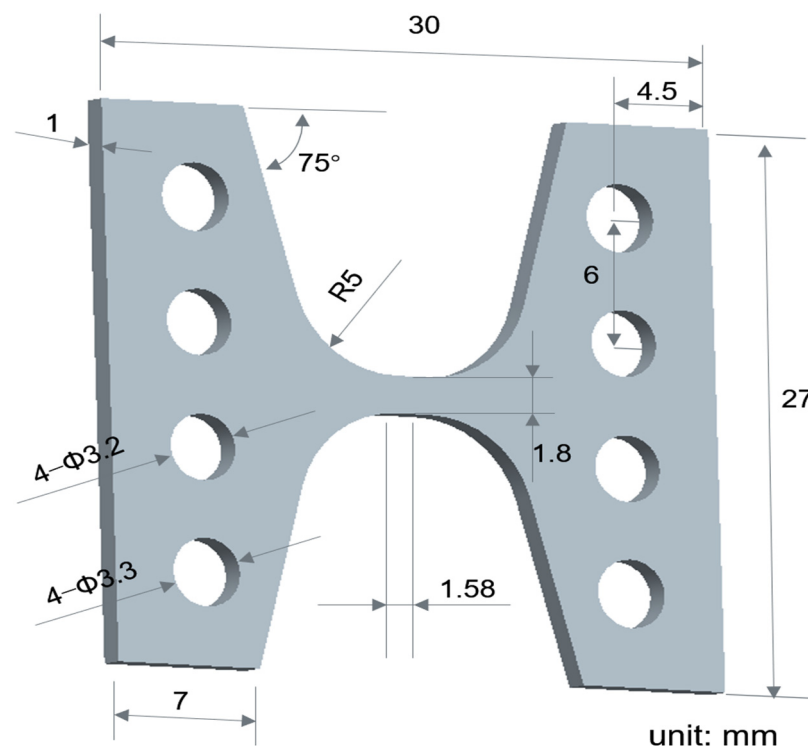


Figure 2. Geometric dimensions of specimen for in situ uniaxial tensile observation.

2.3. Method for Identifying the Activated Slip System in the α_2 Phase

In previous research, the method to identify different slip systems by the tensile test in α/β titanium alloy (Ti-6Al-4V) [28] and near- β titanium alloy (Ti-17) [29] using the EBS D experiment had been reported. The method was used and considered reliable by many researchers. For example, Tan et al. [30] used this method to determine the activated slip system of the α phase in TC-21. Zhang et al. [22] applied this method to identify the activated slip system of B2 and O phases in Ti-22Al-25Nb. Therefore, it is appropriate

to use this method to identify the activated slip system in the α_2 phase (hexagonal close-packed phase based on Ti_3Al) in the Ti-22Al-25Nb alloy. The method is based on Schmid's law, that is, for a single crystal subjected to tension and the tensile axis along a certain crystal direction, the slip system starts to activate when the critical resolved shear stress (CRSS) of the external force reaches a certain critical value in the slip direction of a certain slip plane. For an equivalent set of slip systems, they were subjected to the same CRSS. Therefore, the one with the largest Schmid factor (SF) should be the easiest to activate. In the process of calculating the SF for 4-axis coordinates, it is usually transformed to the corresponding 3-axis coordinates first for the convenience of calculation based on the Equations (1) and (2).

$$[h'k'l'] = [2h + k \quad \sqrt{3}k \quad \sqrt{3}l(a/c)] \quad (1)$$

$$[u'v'w'] = [(2u + v)\sqrt{3}/2 \quad 3v/2 \quad wc/a] \quad (2)$$

The module of the $[h'k'l']$ and $[u'v'w']$ can be expressed by the Equations (3) and (4). Therefore, after normalization of Equations (1) and (2), the 4-axis coordinates in hexagonal are turned into 3-axis coordinates in cubic, which can be expressed by the Equations (5) and (6) [29].

$$d_{hkl} = [(2h + k)^2 + 3k^2 + 3l^2(a/c)^2]^{1/2} \quad (3)$$

$$d_{uvw} = [3(u + v/2)^2 + 9v^2/4 + w^2(c/a)^2]^{1/2} \quad (4)$$

$$[h'k'l'] = [2h + k \quad \sqrt{3}k \quad \sqrt{3}l(a/c)]/d_{hkl} \quad (5)$$

$$[u'v'w'] = [(2u + v)\sqrt{3}/2 \quad 3v/2 \quad wc/a]/d_{uvw} \quad (6)$$

where h, k and l are parameters of slip plane; u, v and w are parameters of slip direction; a, b and c are lattice parameters.

Tensile direction T in sample coordinate should be transformed into the corresponding vector in crystal coordinate T^c by the Equations (7) and (8) [31]. In the present research, the specimens are subjected to tensile force only on the X-axis, so tensile direction $T = [1 \quad 0 \quad 0]^T$ in the sample coordinate.

$$T^c = gT \quad (7)$$

$$g = \begin{Bmatrix} \cos \varphi_1 \cos \varphi_2 - \sin \varphi_1 \sin \varphi_2 \cos \phi & \sin \varphi_1 \cos \varphi_2 + \cos \varphi_1 \sin \varphi_2 \cos \phi & \sin \varphi_2 \sin \phi \\ -\cos \varphi_1 \sin \varphi_2 - \sin \varphi_1 \sin \varphi_2 \cos \phi & -\sin \varphi_1 \cos \varphi_2 + \cos \varphi_1 \sin \varphi_2 \cos \phi & \cos \varphi_2 \sin \phi \\ \sin \varphi_1 \sin \phi & -\cos \varphi_1 \sin \phi & \cos \phi \end{Bmatrix} \quad (8)$$

where φ_1, φ_2 and ϕ are the crystal orientation Euler angles of the α_2 phase in the area of interest (AOI) obtained by the EBSD experiment. After the transformation of the above equation, the SF of the slip system can be expressed by the Equation (9).

$$\text{SF} = (n \cdot T^c)(s \cdot T^c) \quad (9)$$

where n is the normal direction of slip plane, s is the slip direction and T^c is the tensile direction in crystal coordinate.

The slip lines, which are the intersection lines of the slip plane and the specimen surface, are generated when the specimen is continuously subjected to the tensile test. To determine which slip system is most easily to be activated, the theoretical angle (theoretical θ) for every possible activated slip system needs to be calculated. When the theoretical angle is the same or close to the generated angle (which can be measured by the slip line and the tensile direction), the slip system can be considered to be activated. The theoretical θ can be calculated by the Equation (10).

$$\cos \theta = (ge_z \times n) \cdot T^c \quad (10)$$

where e_z is the unit normal vector of sample surface in crystal coordinate, i.e., (001) plane in sample coordinate; n is the normal direction of slip plane; T^c is the corresponding vector in crystal coordinate.

3. Results and Discussion

3.1. Microstructure and In Situ Tensile Observation

3.1.1. Microstructure

The microstructure of the alloy after solution treated at 1020 °C for 30 min is shown in Figure 3. It is a typical equiaxed microstructure, which consists of the equiaxed α_2 particles, B2 matrix and a few needle-like lamellar O phases. Compared with the microstructure after forging shown in Figure 1b, the volume fraction and size of the α_2 phases after solution treatment remain almost unchanged, while the content and thickness of lamellar O phases are drastically decreased and the volume fraction of the B2 matrix is significantly increased. At the same time, the equiaxed α_2 /O particles are transformed into equiaxed α_2 particles. Based on the phase diagram of Ti-22Al-25Nb established by Raghavan [32], the solution treatment temperature of 1020 °C is in the (α_2 + B2) phase region, thus leading to a significant dissolution of the O phases. The presence of a small amount of needle-like lamellar O phases is mainly due to the short solution time, which makes the O lamellae not completely dissolved into the B2 matrix.

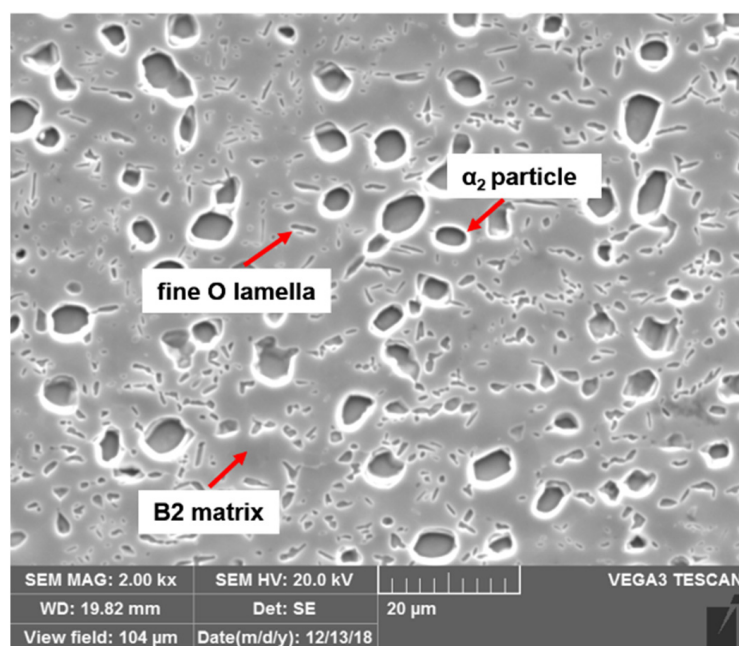


Figure 3. Microstructure of the alloy after solution treatment.

3.1.2. In Situ Tensile Observation

The stress-displacement curve of the solution-treated Ti-22Al-25Nb alloy is shown in Figure 4. Each turning point in the curve represents a short pause for observation and the tensile process will be continued after the observation. As can be observed in Figure 4, there are four stopping points at the elastic deformation stage, and only one stopping point (marked as the first) is recorded because of no significant difference in the microstructure among the four turning points before yielding. From the second stop point to the seventh stop point, the alloy is in the yielding stage. According to the in situ tensile curve in Figure 4, the yield strength (YS) and ultimate tensile strength (UTS) of the alloy with equiaxed α_2 particles are 574 MPa and 601 MPa, respectively. The overall elongation of the alloy is 8.23%. Microstructure observations of the in situ tensile process according to different stages in Figure 4 are shown in Figure 5. The microstructural morphology has undergone a series of major changes during the in situ tensile deformation. At the

elastic deformation stage (Figure 5a), the microstructure had no obvious change from that after solution treatment. When the in situ tensile process reached the second and third stages (Figure 5b,c), a few wide slip bands appeared in the B2 matrix and no slip bands occurred in α_2 particles, indicating that deformation preferentially occurred in the B2 matrix to relieve the stress. At that time, the distance (marked by the blue lines) between the two narrower adjacent slip bands is about 3.8 μm . As the tensile displacement increased (Figure 5d), more inhomogeneous slip bands occurred in the B2 matrix and the former slip bands became more well-defined. When the tensile process reached the fifth stage (Figure 5e), a large number of slip bands including cross-slip bands existed in the B2 matrix and the distance (marked by the blue lines) between two adjacent slip bands was reduced significantly to approximately 1.5 μm . This indicates that continuous plastic deformation occurs in the B2 matrix and the appearance of cross slip bands is beneficial for releasing the stress caused by dislocation piling up [33]. Meanwhile, some of the slip bands shear the smaller α_2 particles, while more slip bands were blocked by α_2 particles, especially the larger ones. This indicated that larger α_2 particles hinder the movement of the slip bands, while smaller α_2 particles have a limited hindering effect on the movement of the slip bands. In other words, while a large number of slip bands appeared in the B2 matrix, only a few appeared in the α_2 particles. It revealed that the plastic deformation of the alloy during the tensile process is mainly contributed to by the B2 phase. That is consistent with the results proposed by Boehlert [12] that the B2 matrix can enhance the plastic properties of the Ti_2AlNb -based alloy compared with the O phase and the α_2 phase. It is known that the B2 phase as a body-centered cubic (BCC) structure phase has three independent slip systems [17], such as $\{110\}\langle 111 \rangle$ slip systems, $\{112\}\langle 111 \rangle$ slip systems and $\{123\}\langle 111 \rangle$ slip systems. Each independent slip system has lots of equivalent slip modes. Therefore, during the tensile deformation of the alloy, there can be more activated slip systems in the B2 phase to relieve the stress concentration. It should be noted that the direction of the slip bands (marked by the red line in Figure 5b) from Figure 5b to Figure 5e appearing in the B2 matrix is basically identical. This is because the deformation occurs within a single B2 grain and only one slip system is activated for dislocation slipping in the B2 grain.

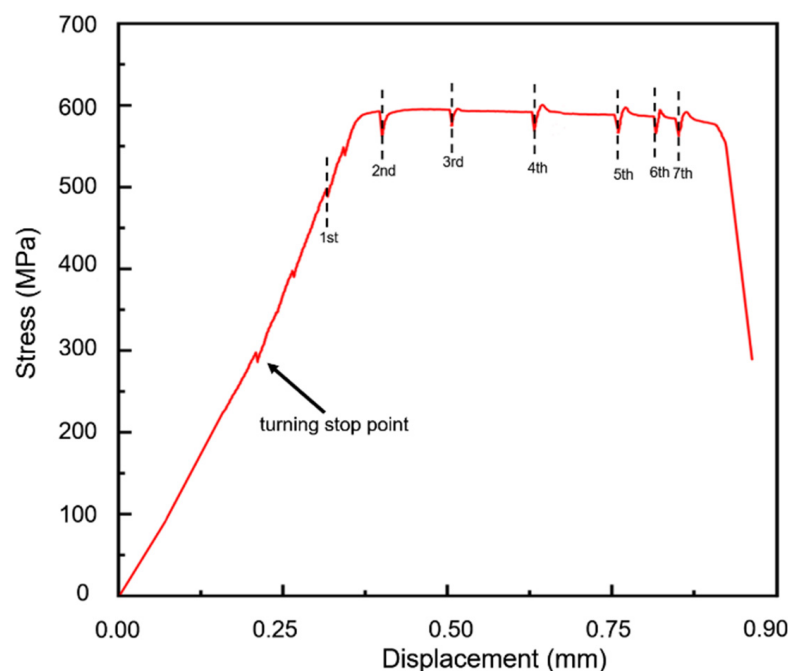


Figure 4. In situ room temperature tensile stress-displacement curve of the solution-treated Ti-22Al-25Nb alloy.

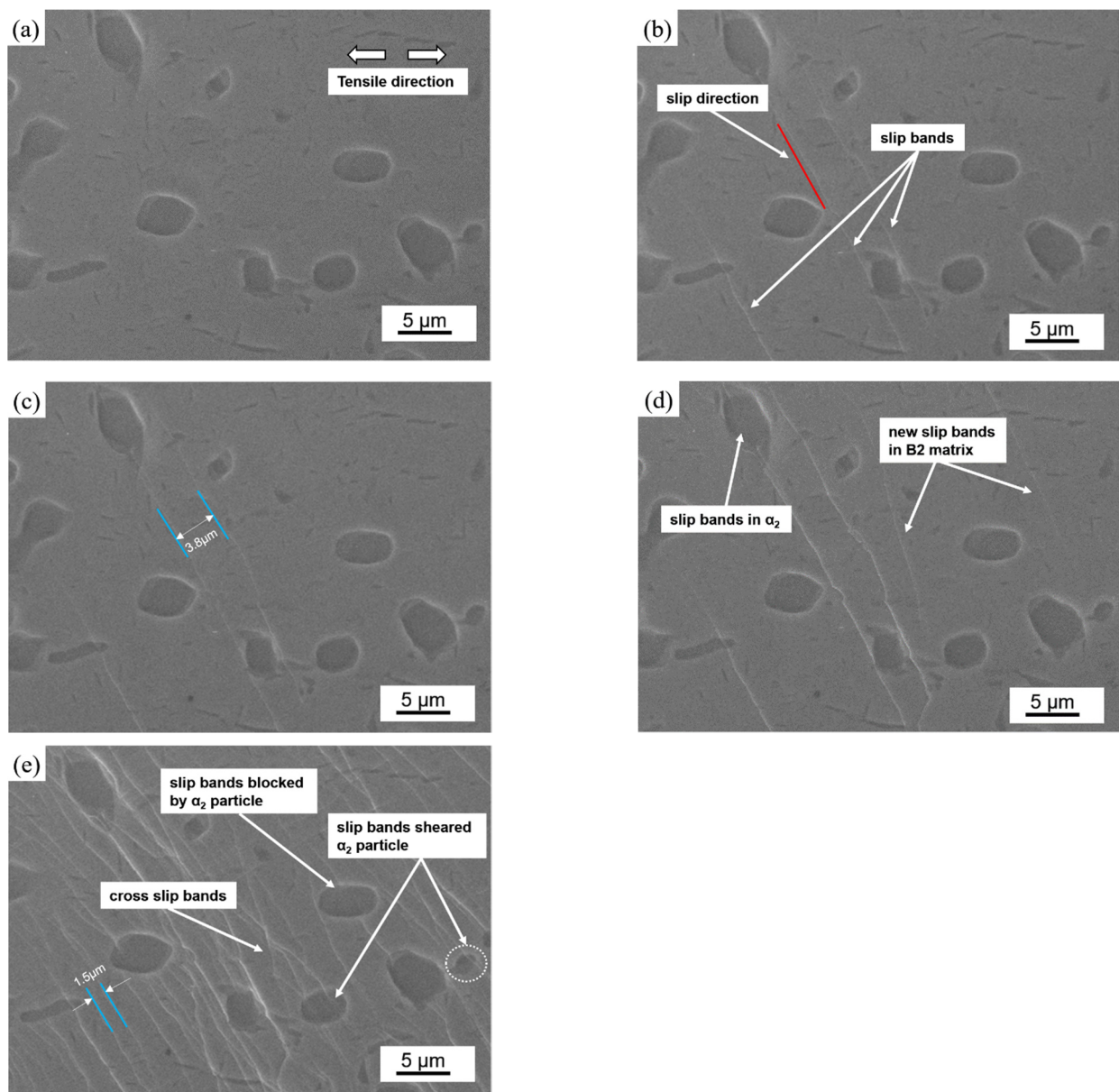


Figure 5. The SEM images during the in situ tensile test according to different stages in Figure 4: (a) 1st, (b) 2nd, (c) 3rd, (d) 4th and (e) 5th.

The morphology of slip bands at different locations of the specimen at the 5th stage of the stress-displacement curve is shown in Figure 6. As can be observed in Figure 6a, most of the slip bands are obstructed by α_2 particles, some of them bypass the particles and only a few can penetrate the particles. It is indicated that the α_2 phases and the B2 phases exist slip incompatibility, and the α_2 phase is a source of the alloy strengthening. Figure 6b shows the movement of the slip bands in different B2 grains. The white dotted line represents the B2 grain boundary and solid lines with different colors (green line, red line and yellow line) represent the direction of the slip bands appearing in different B2 grains. The straight and continuous slip bands in a certain B2 grain are equidistant from each other, indicating that even plastic deformation occurs in a single B2 grain during the tensile process. In #5 B2 grain, there are two groups of slip bands in different directions (marked by yellow lines), indicating that multiple slip bands are activated during deformation. The multiple slip planes ($\{110\}$, $\{112\}$, $\{123\}$) and only one slip direction ($\langle 111 \rangle$) for the B2 phase are the key factors for the appearance of multiple- and cross-slip bands [34]. In different B2 grains (marked by the numbers in Figure 6b), the slip bands have different

directions, indicating that these B2 grains hold different orientations and therefore different slip systems are activated during the uniaxial tensile process. Meanwhile, the slip bands within the B2 grains are obstructed by the B2 grain boundaries, such as the grain boundary between #1 and #2 B2 grain and the grain boundary between #2 and #5 B2 grain. For the grain boundary between #1 and #4 B2 grain, there is only a limited distance of the slip bands passing through the grain boundary. This indicates that not only the α_2 particles, but also the B2 grain boundaries also have a strengthening effect on the alloy. Figure 6c is a partially enlarged view of Figure 6b. It can be observed from Figure 6c that the slip bands penetrate the needle-like lamellar O phase without changing the slip direction. Compared with the α_2 particles (Figure 6a), the hindering effect of O phases on the motion of the slip bands is almost negligible. In other words, the slip transmission between the B2 phase and the needle-like O phases is much easier than that between the B2 phase and the α_2 phase. This is mainly due to the multiple $\langle c+a \rangle$ type slips that were activated even in grain orientations which do not specially favor the slip [26]. Moreover, the activated slip systems in the B2 phase can induce a single and/or double system slip in the O phases [22]. Therefore, the slip compatibility between B2 and the O phases is better than that of the B2 and the α_2 phases, resulting in slip bands shearing across the needle-like O phase and being hindered at the α_2 particles. The crack initiation and propagation behavior during the in situ tensile test will be discussed in detail in a later subsection.

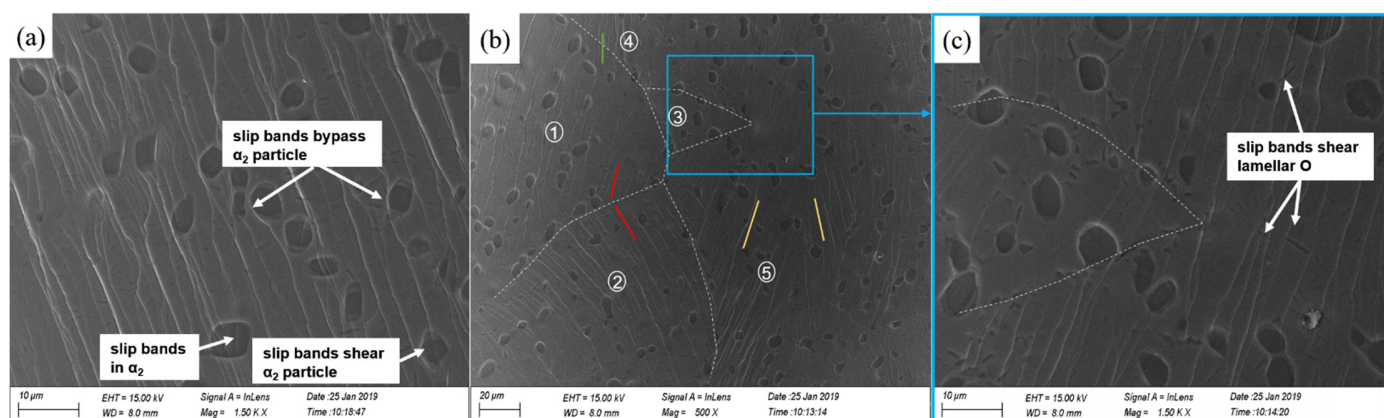


Figure 6. The morphology of slip bands at different locations of the specimen: (a), (b) the 5th stage of the stress-displacement curve; (c) a higher magnification of (b).

3.2. Analysis of Slip System for α_2 Phase

Based on the above analysis, it is evident that α_2 particles have an inhibitory effect on the movement of the slip band, which significantly affects the mechanical properties of the alloy, particularly plasticity. Therefore, in order to better understand the deformation mechanism of the alloy with the equiaxed microstructure, it is essential to analyze the activated slip systems in the α_2 phase. As mentioned in Section 2.2, the specimens were subjected to EBSD experiments prior to the in situ tensile test for obtaining the orientation data (Euler angles) of the α_2 particles. The inverse pole figure (IPF) map of α_2 particles in the Area of Interest (AOI) is shown in Figure 7. The Euler angles (ϕ_1 , ϕ , ϕ_2) were collected for each α_2 particle to analyze the activated slip system. Figure 8a shows the distribution of slip bands at the fifth stage of the stress-displacement curve. The red line represents the direction of the slip bands within α_2 particles and the observed θ represents the angle between the tensile direction and the slip bands. Figure 8b shows the corresponding numbering of the 26 α_2 particles in Figure 8a. From Figure 8a, it is found that the planar slip appears in α_2 particles and only 15 α_2 particles show the slip bands. It should be noted that from this stage to the final fracture, no new slip bands are generated inside the α_2 particles. Using the procedure mentioned in Section 2.3, the determination of the activated slip system was analyzed for the #10 α_2 particle as an example. As shown in Table 2, there is a maximum value of Schmid Factor ($SF_{\max}$) in each of the three different

types of slip systems, which represents the most possible activated slip system in that type. The three most possible activated slip systems are (0001)[$\bar{2}$ 110] for basal $\langle a \rangle$ type, ($\bar{1}$ 100)[11 $\bar{2}$ 0] for prism $\langle a \rangle$ type and ($\bar{2}$ 111)[$\bar{2}$ 1 $\bar{1}$ 6] for pyramidal $\langle a+c \rangle$ type, respectively. By matching the calculated angle (110.2°) to the observed angle (110°), (0001)[$\bar{2}$ 110] is considered the activated slip system in the #10 α_2 particle. The activated slip systems of all α_2 particles in AOI can be identified by this method, as shown in Table 3. There is a small deviation between the observed angle and the calculated angle (less than 4.1°). The distribution of SF derived from EBSD data is shown in Figure 9. In the scale of Figure 9, the lighter the color represents the larger value of SF; that is, the slip system is more likely to be activated with the larger value of SF due to the CRSS being assumed to be identical to the equivalent slip systems. It can be observed that Figure 9 is in accordance with the results in Table 3. Figure 10 shows the number and the SF distribution of the activated slip systems for three different types. It is found that basal $\langle a \rangle$ type, prism $\langle a \rangle$ type and pyramidal $\langle a+c \rangle$ type slip systems are all activated during room temperature deformation. As shown in Figure 10b, three different types of slip systems are activated with a larger SF (above 0.3) and the number of basal $\langle a \rangle$ type and prism $\langle a \rangle$ type slip systems significantly increase with increasing SF. Moreover, the prism $\langle a \rangle$ type slip appears to be more easily activated than the basal $\langle a \rangle$ type slip due to the larger number of activated slip systems and smaller SF. It should be noted that the activated pyramidal $\langle a+c \rangle$ type slip system has the largest of all SF (0.496) and only one of them is activated. This suggests that the pyramidal $\langle a+c \rangle$ type slip system is the hardest slip to be activated among the three slip systems and the conditions for activating the pyramidal $\langle a+c \rangle$ type slip are severe. Therefore, the prism $\langle a \rangle$ type slip is most likely to activate for the α_2 phase during the room temperature tensile process, followed by the basal $\langle a \rangle$ type slip and finally the pyramidal $\langle a+c \rangle$ type slip. A similar phenomenon is also found by Legros et al. [27]. However, the present results are not in agreement with the results in Ti₃Al [23] that only $\langle a \rangle$ type slips are observed in the α_2 phase during room and intermediate temperatures deformation. This is mainly because of the higher niobium (Nb) content in Ti₂AlNb-based alloys than in Ti₃Al alloys. Previous studies [25] suggest that a Ti₃Al alloy with 4 at % Nb decreases the Peierls stresses of slip systems and increases the mobility of both $\langle 11\bar{2}0 \rangle$ and $\langle 11\bar{2}6 \rangle$ dislocations, leading to a higher ductility than a Ti₃Al alloy. Moreover, the addition of Nb elements reduces the difference in CRSS between pyramidal $\langle a+c \rangle$ type slips and prism $\langle a \rangle$ type slips [35], indicating that the anisotropy of deformation is decreased by the addition of Nb elements. For different types of slip systems, i.e., the non-equivalent slip systems, CRSS needs to be taken into account. The CRSS for basal $\langle a \rangle$ type slips is approximately three times that of the prism $\langle a \rangle$ slips in Ti₃Al alloys, while the addition of Nb increases basal glide, resulting in a closer CRSS for basal $\langle a \rangle$ type slip and prism $\langle a \rangle$ type slip [26]. Thus, for basal $\langle a \rangle$ and prism $\langle a \rangle$ type slips with the same number of equivalent slip systems, it can be observed from Figure 10a that the number of activated slip systems between these two types of slip systems is quite similar. As shown in Table 2 and Figure 9, SF_{max} for pyramidal $\langle a+c \rangle$ type slip ($\bar{2}$ 111)[$\bar{2}$ 1 $\bar{1}$ 6] is 0.482, which is much larger than SF_{max} of 0.331 for basal $\langle a \rangle$ type slip (0001)[$\bar{2}$ 110] in #10 α_2 particle. In addition, the angle between the c-axis and the tensile direction is only about 19° , which is very close to the [1] direction, as shown in Figure 11. However, the slip system is not activated for the pyramidal $\langle a+c \rangle$ type slip. In contrast, the basal slip system is activated with the smaller SF, indicating that CRSS for the pyramidal $\langle a+c \rangle$ type slip is much higher than basal $\langle a \rangle$ type slip. Based on an experimental result, the CRSS for pyramidal $\langle a+c \rangle$ type is about seven times higher than that of prism $\langle a \rangle$ slips in a Ti₃Al-Nb alloy [35], even though the addition of Nb reduces the ratio of CRSS in pyramidal $\langle a+c \rangle$ type slips and prism $\langle a \rangle$ type slips. Therefore, the deformation in the α_2 phase is mainly contributed by the prism $\langle a \rangle$ type slip and the basal $\langle a \rangle$ type slip.

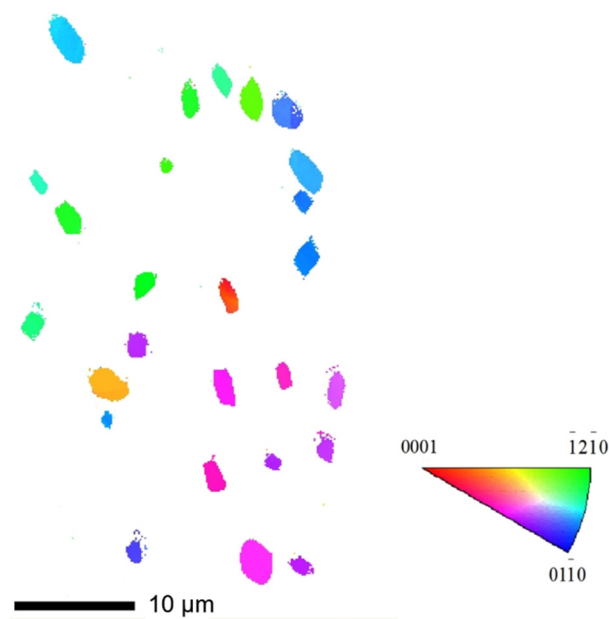


Figure 7. IPF map of α_2 particles in the Area of Interest (AOI).

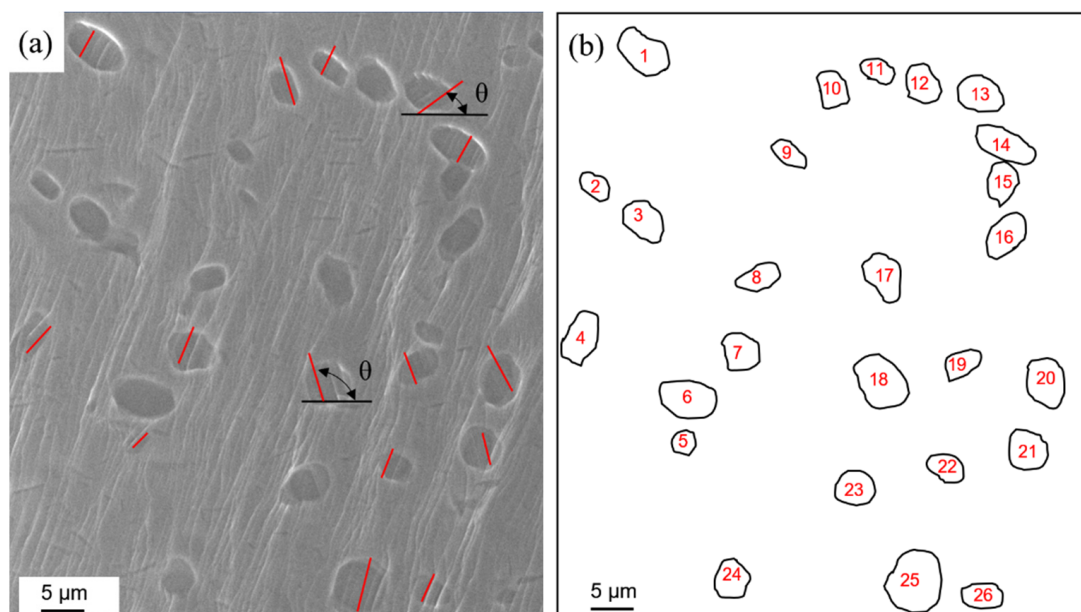


Figure 8. (a) Slip bands and observed θ of α_2 particles in the AOI; (b) schematic diagram of particle position.

From Figure 8a, it is also found that the slip band in the B2 phase passed directly through a small number of α_2 particles such as #22 α_2 particle and #26 α_2 particle without deflection. This indicates that there is a good slip compatibility between these two α_2 particles and the B2 phase. The pole figures of #22 α_2 particle and #26 α_2 particle with the B2 phase pole figure are shown in Figures 12 and 13, respectively. The $\{0001\}$ plane of the #22 α_2 particle is parallel to $\{110\}$ plane of the B2 phase as marked by the red ellipse in Figure 12, and $\langle 11\bar{2}0 \rangle$ direction of the #22 α_2 particle is also parallel to $\langle 111 \rangle$ direction of the B2 phase as marked by the dark green triangle in Figure 12. Thus, the #22 α_2 particle maintained the specific orientation relationship, i.e., $\{0001\}\alpha_2 // \{110\}\text{B2}$ and $\langle 11\bar{2}0 \rangle \alpha_2 // \langle 111 \rangle \text{B2}$. Similar to the #22 α_2 particle, the #26 α_2 particle also maintains specific orientation relationships with the B2 phase marked by the blue ellipse and purple triangle in Figure 13. Therefore, it can be derived that the Burgers orientation relationships

(BOR) between B2 and α_2 phases is favorable for the slip compatibility. In other words, when the α_2 particles maintain a specific orientation relationship with the B2 matrix, the hindrance of the α_2 particles to the movement of the slip bands is significantly reduced.

Table 2. The analysis of the activated slip system for #10 α_2 grain.

Slip Type	Slip Plane (hkl)	Slip Direction [uvw]	SF	Calculated θ (°)
Basal slip <a> type	0001	11 $\bar{2}$ 0	0.211	69.8/110.2
		1 $\bar{2}$ 10	0.121	
		2 $\bar{1}$ 10	0.331	
Prism slip <a> type	1 $\bar{1}$ 00	11 $\bar{2}$ 0	-SF _{max} 0.063	29.1/150.9
		1 $\bar{2}$ 10	-SF _{max} 0.043	
		2 $\bar{1}$ 10	0.019	
Pyramidal slip <a+c> type	11 $\bar{2}$ 1	1 $\bar{1}$ 26	0.061	41.9/138.1
		11 $\bar{2}$ 6	0.405	
		1 $\bar{2}$ 16	0.145	
		1 $\bar{2}$ 11	0.342	
		2 $\bar{1}$ 16	0.482	
		2 $\bar{1}$ 16	-SF _{max} 0.330	

Table 3. The activated slip system of α_2 particles in the AIO.

No.	Observed θ (°)	Calculated θ (°)	SF	Activated Slip System	Slip Type
1	59	57.3	0.312	(1 $\bar{1}$ 0) [11 $\bar{2}$ 0]	Prism <a> type
2	-	-		No slip system activated	
3	-	-		No slip system activated	
4	45	47.4	0.455	(0001) [2 $\bar{1}$ 1 0]	Basal <a> type
5	50	52.3	0.473	(0001) [2 $\bar{1}$ 1 0]	Basal <a> type
6	-	-		No slip system activated	
7	70	70.4	0.463	(10 $\bar{1}$ 0) [1 $\bar{2}$ 1 0]	Prism <a> type
8	-	-		No slip system activated	
9	-	-		No slip system activated	
10	110	110.2	0.331	(0001) [2 $\bar{1}$ 1 0]	Basal <a> type
11	70	71.2	0.453	(0001) [1 $\bar{2}$ 1 0]	Basal <a> type
12	-	-		No slip system activated	
13	41	45.1	0.496	(2 $\bar{1}$ 11) [2 $\bar{1}$ 16]	Pyramidal <a+c> type
14	65	62.8	0.478	(0001) [2 $\bar{1}$ 1 0]	Basal <a> type
15	-	-		No slip system activated	
16	-	-		No slip system activated	
17	-	-		No slip system activated	
18	110	114.1	0.473	(10 $\bar{1}$ 0) [1 $\bar{2}$ 1 0]	Prism <a> type
19	109	111.5	0.437	(1 $\bar{1}$ 0 0) [11 $\bar{2}$ 0]	Prism <a> type
20	117	112.6	0.467	(10 $\bar{1}$ 0) [1 $\bar{2}$ 1 0]	Prism <a> type
21	104	104.9	0.443	(10 $\bar{1}$ 0) [1 $\bar{2}$ 1 0]	Prism <a> type
22	70	66.3	0.472	(1 $\bar{1}$ 0 0) [11 $\bar{2}$ 0]	Prism <a> type
23	-	-		No slip system activated	
24	-	-		No slip system activated	
25	76	74.2	0.468	(0001) [1 $\bar{2}$ 1 0]	Basal <a> type
26	70	69.3	0.458	(1 $\bar{1}$ 0 0) [11 $\bar{2}$ 0]	Prism <a> type

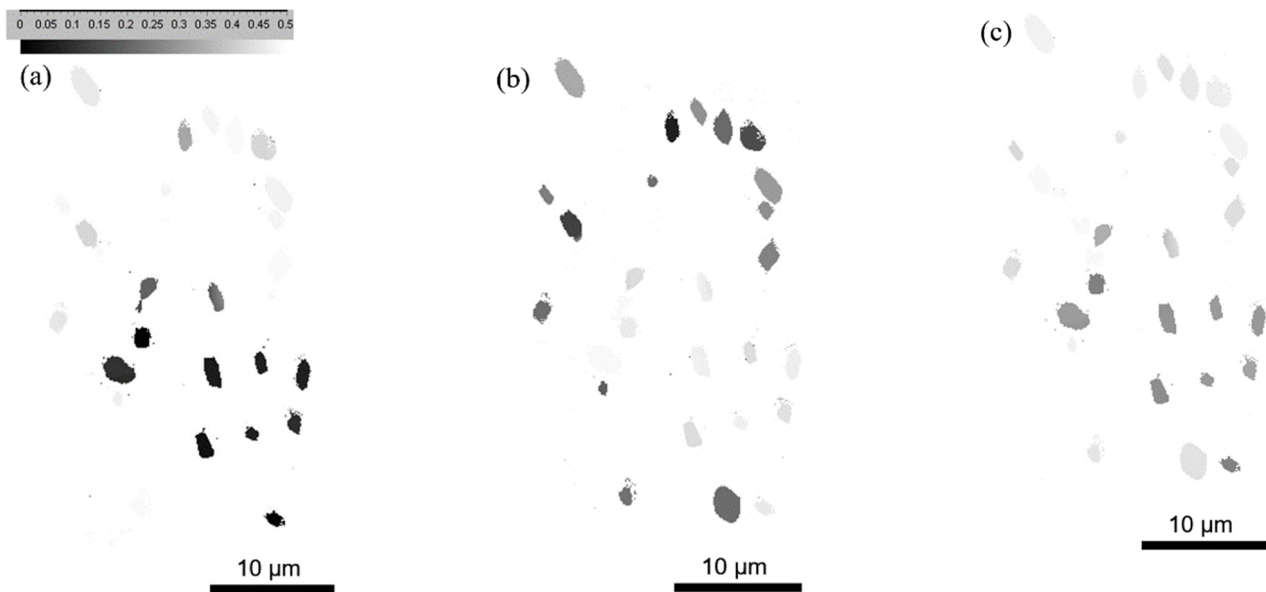


Figure 9. The distribution of SF for different types of slip systems (a) basal $\langle a \rangle$ type, (b) prism $\langle a \rangle$ type and (c) pyramidal $\langle a+c \rangle$ type.

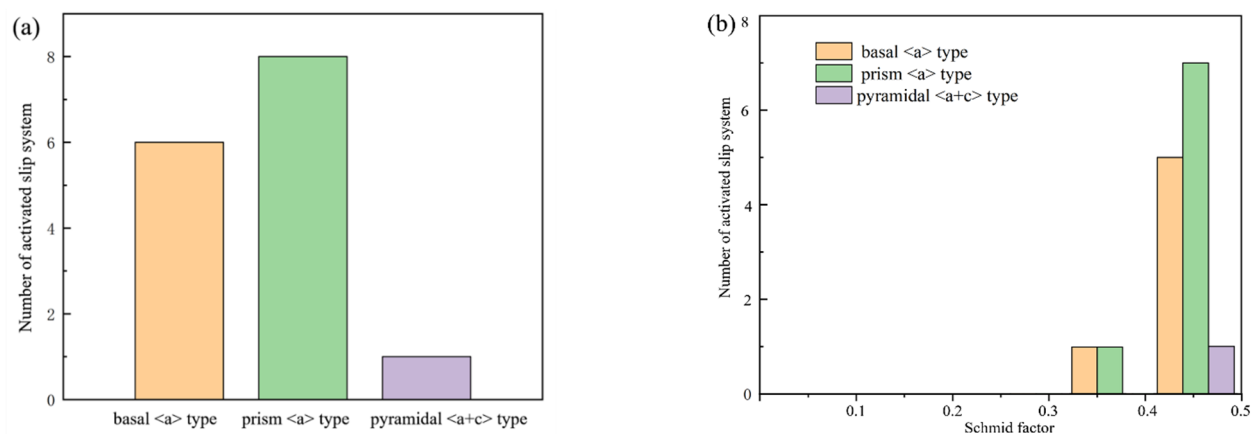


Figure 10. The activated slip systems of α_2 particles (a) number of slip systems; (b) distribution of SF in different slip systems.

3.3. Crack Initiation and Propagation

Figures 14 and 15 show the process of crack nucleation and propagation of Ti-22Al-25Nb during the room temperature deformation. As shown in Figure 14a, a few microcracks are generated at the α_2 /B2 phase boundaries in the 5th stage of the stress-displacement curve. The area marked by the purple ellipse and the orange rectangle in Figure 14 is where we focus our attention on the continuous tensile process. As the tensile process continues, microcracks nucleate not only at the α_2 /B2 phase boundaries (marked by the red and black arrows in Figure 14) but also within the α_2 phase (Figure 15a). With the further increase in tensile displacement, the microcracks propagate along the α_2 /B2 phase boundaries (Figure 14c) and/or across the α_2 particles (Figure 15b). However, no microcracks are initiated in the B2 matrix due to the multiple activated slip systems of the B2 phase, which also indicates that the overall plastic deformation of the alloy is controlled by the B2 phase. A similar phenomenon was found by Wu et al. [36] in a study of the tensile deformation of Ti-24Al-14Nb-3V-0.5Mo alloy. Therefore, the α_2 /B2 phase boundaries and the interior of α_2 particles are the favored sites for microcracks' formation. According to the analysis in the previous section, the deformation of the α_2 phase is mainly concentrated only in

prism $\langle a \rangle$ type slip and basal $\langle a \rangle$ type slip systems during the room temperature tensile deformation. Moreover, only three equivalent slip modes are included for each type of slip system. Therefore, it is difficult for the α_2 phase to release the stress through deformation due to a few activated slip systems, which leads to the nucleation of microcracks within the α_2 particles. In addition, the different crystal structures of the α_2 and B2 phases make the stress tend to be concentrated at the α_2 /B2 phase boundary, leading to the nucleation of microcracks at the α_2 /B2 phase boundary. As can be observed in Figure 15, there is a lot of crack propagation within the α_2 particle that is not fully developed. When a crack is initiated on the basal plane and activates co-planar $\langle a \rangle$ slip, large normal stresses near the crack tip are generated. The stress can only be relieved when the $\langle a+c \rangle$ slip system is activated [23]. However, the $\langle a+c \rangle$ slip system is difficult to be activated due to the larger CRSS and specific orientation relationship with the c-axis. Therefore, the cracks continue to grow within the α_2 particle. When the crack moves into the B2 matrix, the propagation of the crack is not fully developed because the B2 phase is beneficial to the blunt of the microcracks [12]. According to the above analysis, the schematic diagram of the initiation and propagation of the crack is shown in Figure 16. Based on previous discussions, the Ti-22Al-25Nb alloy with the addition of more Nb elements did not lead to the initiation of more types of slip systems in the α_2 phase compared to the Ti₃Al-Nb alloy. Due to seldom activated slip systems, most of the slip bands in the B2 matrix are obstructed by α_2 particles, except for those that maintain a specific orientation relationship with the B2 matrix. Hall [37] pointed out that the length of the slip band is proportional to the grain size. Therefore, the presence of α_2 particles leads to a decrease in the mean free path of slip (the size of B2 grain) in the B2 phase. According to the Hall–Petch Equation (11), the relationship between grain size and yield strength (YS) is expressed as follows [38]:

$$\sigma_y = \sigma_0 + k_{gb}d^{-1/2} \quad (11)$$

where σ_0 is the lattice friction stress against the movement of a single dislocation, k_{gb} is the constant related to the material properties, σ_y is the yield strength, and d is the average diameter of the grain. Therefore, in the case of a two-phase Ti-22Al-25Nb alloy with equiaxed α_2 particles, the yield strength of the alloy is increased due to the fact that the α_2 particles reduce the mean free path of the slip (the size of B2 grain). In general, the B2 phase provides the main plastic deformation and the α_2 phase is the main strengthening phase of the alloy. Therefore, in order to improve the plasticity of the two-phase alloy with equiaxed α_2 particles, we can reduce the volume fraction of α_2 particles and keep the α_2 particles in a specific orientation relationship with the B2 matrix. On the other hand, we can increase the volume fraction of α_2 particles and break the specific orientation relationship of particles with the B2 matrix to enhance the alloy strength. The fracture morphology of the alloy after the in situ room temperature tensile test is shown in Figure 17. The fracture surface shows the quasi-cleavage feature that contains a large number of small and shallow dimples on planar facets, indicating good ductility of the alloy. The shallow dimples indicate that the fracture originated at the α_2 /B2 phase boundary where microvoids initiated. The growth and the coalescence of the microvoids resulted in the propagation of microcracks.

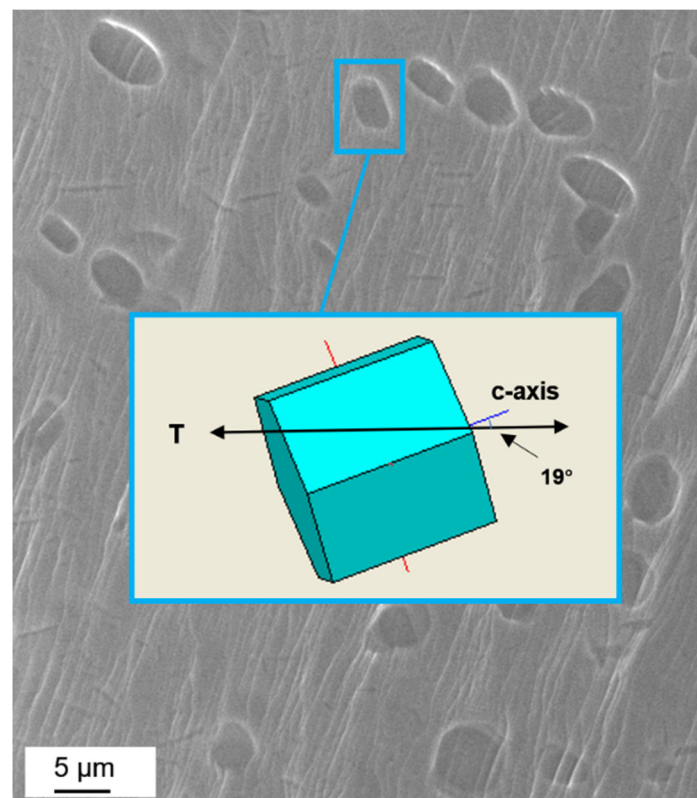


Figure 11. The crystallographic orientation of #10 α_2 grain.

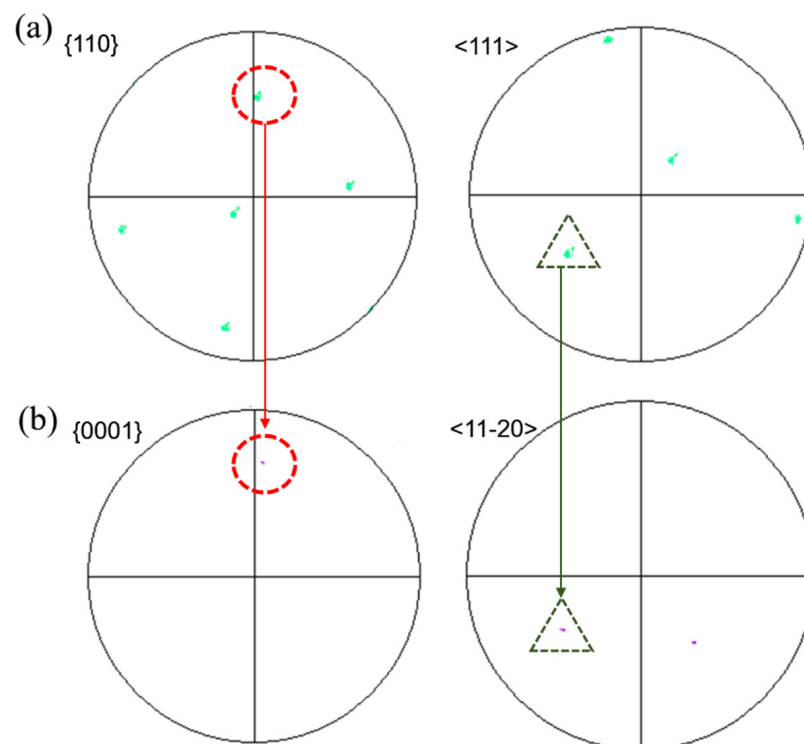


Figure 12. (a) {110} and <111> pole figures of the B2 phase, (b) {0001} and <11 $\bar{2}$ 0> pole figures of the #22 α_2 particle.

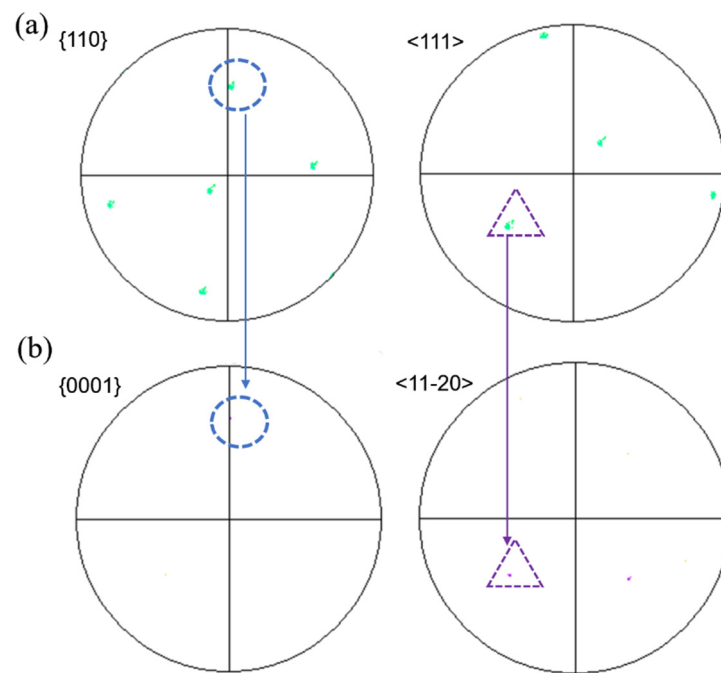


Figure 13. (a) $\{110\}$ and $\langle 111 \rangle$ pole figures of the B2 phase, (b) $\{0001\}$ and $\langle 11\bar{2}0 \rangle$ pole figures of the #26 α_2 particle.

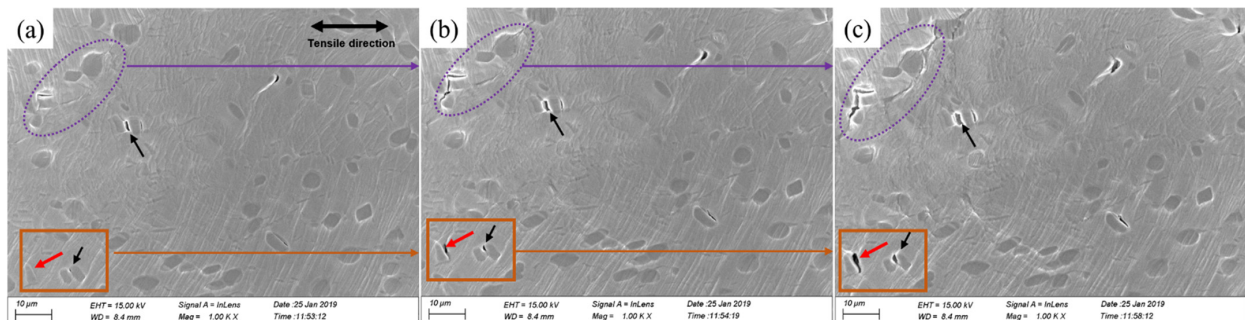


Figure 14. The SEM images during in situ tensile test according to different stages of Figure 4: (a) 5th, (b) 6th and (c) 7th.

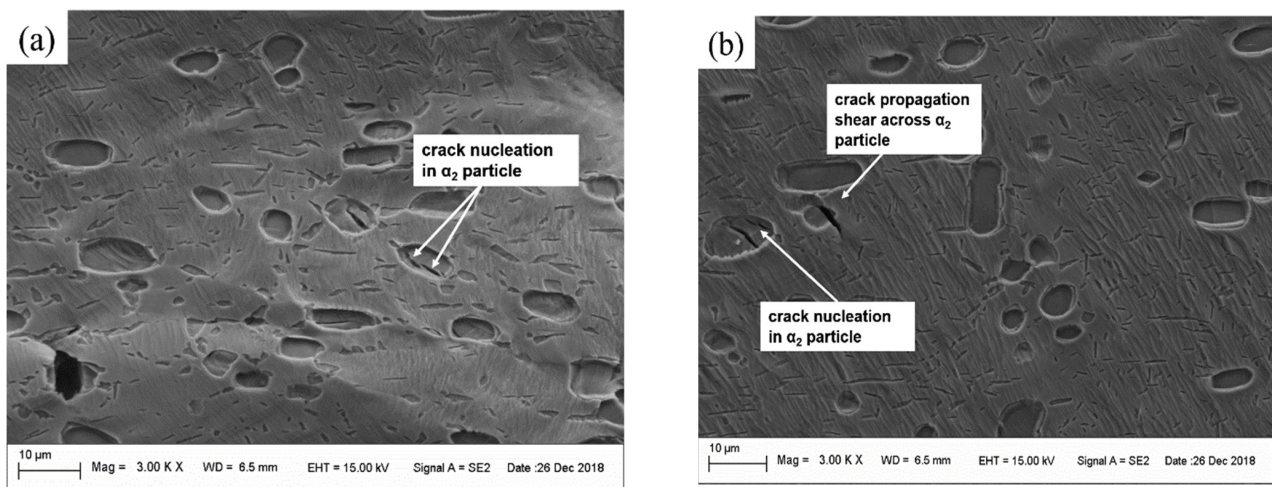


Figure 15. The SEM micrographs of (a) crack nucleation within the α_2 particle; (b) crack propagation shear across the α_2 particle.

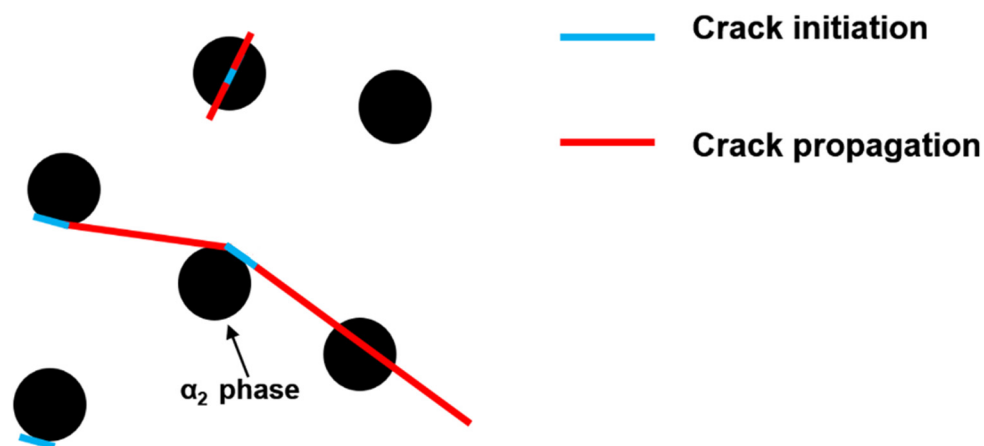


Figure 16. The schematic diagram of the initiation and propagation of the crack [36].

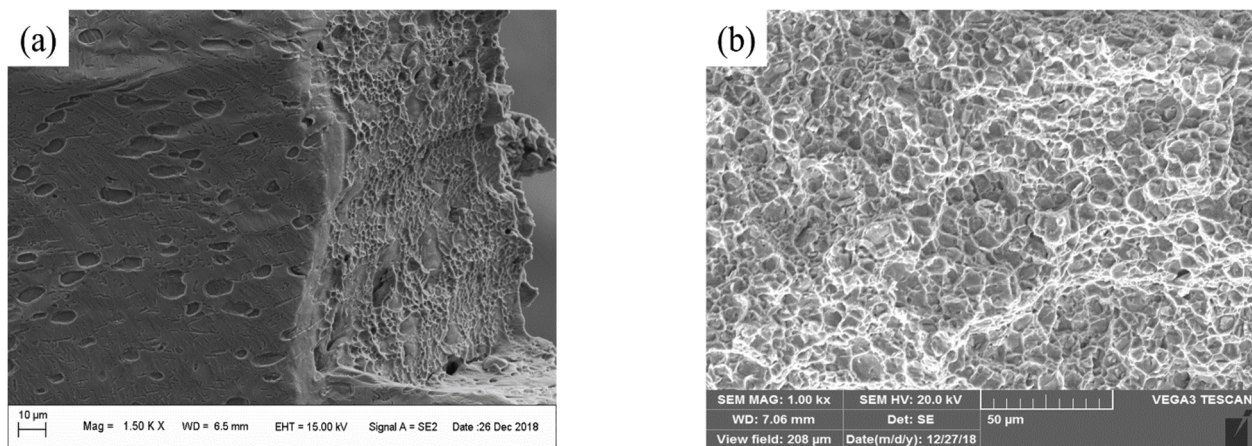


Figure 17. The SEM micrographs of the fracture morphology: (a) side view; (b) front view.

4. Conclusions

In the present work, the characterization of tensile deformation of Ti-22Al-25Nb alloy with equiaxed α_2 particles was investigated through the in situ tensile test. The activated slip system and deformation characteristics of α_2 particles at room temperature were also examined by a combination of in situ tensile tests and EBSD experiments. The main conclusions of this work are as follows:

1. With the tensile displacement increases, a large number of slip bands including multiple- and cross-slip bands appear in the B2 matrix and the distance between two adjacent slip bands decreases significantly. Meanwhile, the movement of the slip bands is hindered by the α_2 particles and the B2 grain boundaries. From the beginning of the tensile process to the final fracture, there are lots of α_2 particles without slip bands.
2. The slip bands penetrate the needle-like lamellar O phase without changing the slip direction. Compared with the α_2 particles, the hindering effect of needle-like O phases on the motion of the slip bands is quite small.
3. For room temperature tensile deformation of α_2 phase, there are three types of slip systems that can be activated, including the prism $\langle a \rangle$ type slip, the basal $\langle a \rangle$ type slip and the pyramidal $\langle a+c \rangle$ type slip. The prism $\langle a \rangle$ type slip is most likely to be activated, followed by the basal $\langle a \rangle$ type slip and finally the pyramidal $\langle a+c \rangle$ type slip. The critical resolved shear stress (CRSS) for the pyramidal $\langle a+c \rangle$ type slip is the highest among the three types of slip systems.

4. The microcracks nucleated at the α_2 /B2 phase boundaries or within the α_2 particles, and microcracks propagated along the α_2 /B2 phase boundaries or across the α_2 particles. The fracture surface shows the quasi-cleavage feature, which contains a large number of small and shallow dimples on planar facets.

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Data Availability Statement: The data are not publicly available due to the fact that these data are part of an ongoing study.

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