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Investigation on microstructure and texture evolution of Ti–6Al–3Nb–2Zr–1Mo alloy during hot deformation

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Abstract

Hot compression tests of Ti–6Al–3Nb–2Zr–1Mo alloy were conducted in the temperature range of 900 °C–1100 °C and strain rate range of 0.01–1s⁻¹. Based on the true stress – true strain curve, the calculated activation energy Q in α+β two-phase region and single-β phase region are 605.85 and 132.44 kJ mol⁻¹, respectively. The microstructure and texture evolution were analyzed by using EBSD technique. The continuous dynamic recrystallization (CDRX) and discontinuous dynamic recrystallization (DDRX) mechanisms are the two dynamic recrystallization (DRX) mechanisms of Ti–6Al–3Nb–2Zr–1Mo alloy deformed at 900 °C, and the latter is dominant. With the increase of temperature to 980 and 1020 °C, CDRX gradually weakens, and the DRX mechanism changes to be controlled by DDRX. The orientation distribution function (ODF) maps show that the initial texture gradually vanishes, and different textures are formed during hot deformation. DDRX behavior causes the decrease of the texture density of Ti–6Al–3Nb–2Zr–1Mo alloy deformed at 900 °C. However, the parallel precipitation of α laths rapidly increases the texture density of Ti–6Al–3Nb–2Zr–1Mo alloy deformed at 980 and 1020 °C.

1. Introduction

Titanium alloys have a wide application in marine engineering owing to the low density, high strength, high toughness, and good corrosion resistance [1–4]. High strength near α titanium alloys are considered to be suitable metallic materials for applications in marine engineering. For instance, the pressure-resistant hulls of submersibles were recently constructed by high strength near α titanium alloys [5]. In order to solve the poor room temperature formability and meet the high requirements for its mechanical properties, titanium alloys usually undergo thermomechanical processing, during which precise shapes and good microstructures can be obtained [6, 7]. High strength near α titanium alloys exhibit complex deformation mechanisms and microstructure evolution during thermo-mechanical processing. Work hardening, dynamic recrystallization (DRX), dynamic recovery (DRV), lamellar globalization, and texture evolution are greatly affected by the deformation parameters and may lead to materials with different ultimate microstructures and mechanical properties [8, 9]. Therefore, it is of great significance to study the microstructure and texture evolution during the hot deformation for optimizing forming process parameters and intensifying the comprehensive properties of titanium alloys.

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In recent years, much effort has been made to study the microstructure and texture evolution of titanium alloys during hot deformation. The activation energy Q is the significant parameter in the process of thermal deformation [10], which can be calculated by the true stress-true strain curve. Zhao et al [11–13] reported that the initial microstructure, phase transition and DRX had significant effects on the deformation activation energy. DRX is a significant feature during the hot deformation of titanium alloys. Ning et al [14] investigated the competition between DRX and DRV during the isothermal deformation of TC18 alloy and demonstrated the dependence of dynamic softening mechanism by characterization of microstructure and power dissipation efficiency. Li et al [15] proposed that DRX was the main softening mechanism of Ti–6.0Al–1.21Nb–9.04Zr–3.88Sn–1.59W–0.28Si alloy, and the average size of DRX grains increased with the decrease of flow stress, and DRX weakened or eliminates the initial texture. S M Abbasi et al [16] proposed that Ti–6V–6Mo–6Fe–3Al alloy exhibited different softening mechanisms under different processing parameters. He et al [17] studied the effect of thermal deformation parameters on the microstructure evolution of Ti–6Al–2Zr–1Mo–1V alloy, discussed the effect of deformation temperature on the lamellae spheroidization mechanism, and summarized the DRX mechanism during the α+β two-phase thermal deformation process. Texture evolution is crucial for the hot deformation of titanium alloy. Li et al [18] found that the transformed α phase showed strong prismatic texture, which is different with the basal texture of compressed α phase. Wu et al [8] found that the texture density of Ti55531 high strength titanium alloy decreased with strain during initial stages, but it increased sharply with further deformation.

Based on the above studies, it can be concluded that the titanium alloy has complex deformation mechanisms during hot deformation. And, different titanium alloys have different DRX mechanisms. Moreover, the study on texture of titanium alloys during hot deformation needs to be continued. Ti–6Al–3Nb–2Zr–1Mo is a typical near-α titanium alloy which has been recently developed by adjusting the type and content of element based on TA15 (Ti–6Al–2Zr–1Mo–1V) titanium alloy for improving its impact toughness. Compared to conventional titanium alloys, Ti–6Al–3Nb–2Zr–1Mo alloy has excellent corrosion resistance and high toughness. The hot working process parameters always have significant effects on microstructure evolution of titanium alloys [19]. However, only a few studies have reported the effect of hot working parameters on the microstructure of Ti–6Al–3Nb–2Zr–1Mo titanium alloy during hot deformation, especially for DRX mechanism and texture evolution.

The purpose of this work is therefore to investigate the microstructure evolution of Ti–6Al–3Nb–2Zr–1Mo alloy during hot deformation. The DRX mechanism and texture evolution are the focus of attention. In this work, the hot compression tests were carried out in the temperature range of 900°C–1100°C and strain rate range of 0.01–1s−1. The flow stress and the calculated activation energy were analyzed. Besides, the influence of temperature and strain rate on the DRX mechanism and texture evolution were discussed. This work could contribute to optimizing the thermal processing parameters, predicting and controlling the hot deformed structure, and improving the product performance of Ti–6Al–3Nb–2Zr–1Mo alloy.

2. Experimental materials and procedures

The Ti–6Al–3Nb–2Zr–1Mo alloy was used in this investigation. It was obtained by vacuum arc melting technology and high temperature forging process. In order to obtain bimodal structure Ti–6Al–3Nb–2Zr–1Mo alloy, the sample was annealed at 980°C for 1h and then cooled to room temperature in air. The true chemical composition of Ti–6Al–3Nb–2Zr–1Mo alloy is shown in Table 1. Its β transus temperature is approximately 998°C [5].

In order to analyze the microstructure and texture evolution of Ti–6Al–3Nb–2Zr–1Mo alloy with bimodal structure during hot deformation. The samples were cut into Ø8 × 12 mm by wire-electrode cutting technique. The hot compression tests were administered in the temperature range of 900°C–1100°C and strain rate range of 0.01–1s−1 on the Gleeble-1500D thermo-mechanical simulator. The flow chart of the hot compression test is shown in figure 1(a). The samples were heated at 10°Cs−1 to deformation temperature and were commanded for 180s for temperature homogenization. Then the samples were compressed into a height reduction of 50%. Finally, water quenching treatment was used to obtain the high temperature deformed microstructure.

The compressed samples were cut along the compression axis, and the mid-section of the cut samples was selected for observation of microstructure and texture. The schematic diagram of the observation position is
shown in figure 1(b). The samples were grinded from 200 mesh to 2000 mesh firstly with sandpaper, and the specimens were electrolytically polished at a voltage of 50 V. The EBSD experiment was performed by using a JSM-7800F field emission scanning electron microscope equipped with an EBSD system. Finally, the EBSD results were analyzed by using the HKL Channel 5 Soft.

3. Results

3.1. Flow stress

Figure 2 shows the true stress-true strain curves of Ti–6Al–3Nb–2Zr–1Mo alloy compressed at different temperatures and strain rates. There are three typical characteristics of metal materials during thermal deformation: work hardening, DRV and DRX [14]. In the early stage of hot deformation, the stress increases sharply, and the curves show work hardening owing to dislocations stacking. As the compression deformation
increases, DRV and DRX begin to appear due to the sufficient deformation energy storage. Thus, the stress increases slowly and begins to decrease after reaching the peak. Finally, the curve achieves dynamic balance between work hardening and dynamic softening, and the stress is in a steady flowing state.

As shown in figure 2(c), the flow stress decreases with increasing deformation temperature. At the strain rate of 1 s\(^{-1}\), the peak stress of the Ti–6Al–3Nb–2Zr–1Mo alloy decreases from 189 MPa to 56 MPa as the temperature increases from 900 to 1100 °C. The α → β phase transformation gradually strengthens with the increasing deformation temperature. The α phase is hexagonal closepacked (HCP) crystal structure of with 4 independent slip systems, and β phases are body-centered cubic (BCC) crystal structure of with 12 independent slip systems, the resistance of deformation decreases with the increase of β phase content. And, the diffusion capability of atoms increases with the increase of deformation temperature, which cause a decrease of flow stress [20]. In addition, the flow stress increases with the increasing strain rate, as shown in figure 2(d), the peak stress of the Ti–6Al–3Nb–2Zr–1Mo alloy increases from 99 MPa to 189 MPa with the strain rate increases from 0.01 to 1 s\(^{-1}\) at 900 °C. At a strain rate of 1s\(^{-1}\), a large number of dislocations are too late to migrate, resulting in the delay of DRX, which leads to high flow stress.

### 3.2. Kinetic analysis

The hot deformation behavior of metallic materials is mainly controlled by activation energy [21]. Generally, the correlation among the deformation temperature, strain rate and peak stress of flow stress during thermal deformation can be expressed as follows [22, 23]:

\[
\dot{\varepsilon} = A\left[\sinh\left(\alpha\sigma\right)\right]^n \exp\left(-\frac{Q}{RT}\right) \quad \text{(for all stress)}
\]

(1)

According to the condition of stress, the hyperbolic sine function in the Arrhenius equation can be presented by the subsequent equations:

\[
\dot{\varepsilon} = A_1\exp\left(-\frac{Q}{RT}\right) \quad \left(\sigma < 0.8\right)
\]

(2)

\[
\dot{\varepsilon} = A_2\exp\left(\beta\sigma\right)\exp\left(-\frac{Q}{RT}\right) \quad \left(\alpha\sigma > 1.2\right)
\]

(3)

where \(\dot{\varepsilon}\) is the strain rate(s\(^{-1}\)), \(\sigma\) is the peak stress (MPa), \(A_1, A_2, n_1\), and \(n_1\) are material constants, \(n\) is the stress exponent, \(Q\) is the activation energy (kJ/mol), \(R\) is the gas constant and \(T\) is the deformation temperature (K). In order to obtain the value of \(n\) and \(Q\), the Equations (1)–(3) can be expressed as:

\[
\ln\dot{\varepsilon} = n\ln\left[\sinh\left(\alpha\sigma\right)\right] + \ln A - \frac{Q}{RT}
\]

(4)

\[
\ln\dot{\varepsilon} = n_1\ln\sigma + \ln A_1 - \frac{Q}{RT}
\]

(5)

\[
\ln\dot{\varepsilon} = \beta\sigma + \ln A_2 - \frac{Q}{RT}
\]

(6)

The \(Q\) can be defined from Equation (7):

\[
Q = R \left[\frac{\frac{d\ln\dot{\varepsilon}}{d\left[\frac{\ln\left[\sinh\left(\alpha\sigma\right]\right]}{d\left(1/T\right)}\right]}}{\frac{d\ln\left[\sinh\left(\alpha\sigma\right)\right]}{d\left(1/T\right)}}\right] \dot{\varepsilon}
\]

(7)

Figure 3 shows the relationships of \(\ln\dot{\varepsilon} = -\ln[\sinh(\alpha\sigma)]\) and \(\ln[\sinh(\alpha\sigma)] - 1000/T\). The values of \(n\) and \(Q\) can be calculated by the slopes of \(\ln -\ln[\sinh(\alpha\sigma)]\) and the slopes of \(\ln[\sinh(\alpha\sigma)] - 1000/T\), and the calculation results are shown in table 2.

### 3.3. Microstructure evolution

#### 3.3.1. Initial microstructure

Figure 4 shows the initial microstructure of Ti–6Al–3Nb–2Zr–1Mo alloy characterized by EBSD technique. The inverse pole figure and grain size distribution are shown in figures 4(a) and (b), respectively. Obviously, the initial microstructure is mainly composed of equiaxed grains and lath grain. The initial average grain size is approximately 6.08 μm. The grain boundary maps and misorientation angle distributions are shown in figures 4(c) and (d), respectively. The low angle grain boundaries (LAGBs, 3° < misorientation angle < 10°), medium angle grain boundaries (MAGBs, 10° ≤ misorientation angle ≤ 15°) and high angle grain boundaries (HAGBs, 15° < misorientation angle) are defined. The maroon, green and navy-blue lines represent the LAGBs, MAGBs and HAGBs, respectively. The average misorientation angle is estimated to be 35.92°, and the fraction of HAGBs reaches up to 58.36%, accounting for a high proportion. However, the distribution of misorientation angles shows a considerable concentration in the range of less than 5°, which is related to the stored dislocation. The dislocation density change and distribution can be obtained from the Kernel Average Misorientation (KAM) maps [24]. Figure 4(c) shows the KAM map of the initial sample, the average value of KAM is 0.34 (figure 4(f)), which indicates that some dislocations formed during the high temperature forging process still remain within the microstructure after the annealing treatment. In addition, LAGBs are mainly distributed in the area of high dislocation density, as is shown by the red line in figures 4(c) and (e).
3.3.2. Grain morphology and size evolution

The inverse pole figures and grain size distributions at 900 °C, are shown in figures 5(a1-c1) and (a2-c2), respectively. A large number of new grains are distributed along the grain boundaries of deformed grains, as shown by the white arrow, which is a typical feature of DRX \[17\]. DRX has a significant influence on the grain size and shape of titanium alloy. In addition, it can be seen from figures 5(a2), (b2) and (c2), the average grain size is 1.61, 1.52 and 1.44 μm at the strain rate of 0.01, 0.1 and 1 s\(^{-1}\), respectively. It is obvious that the grain size after hot compression is refined. This phenomenon can be attributed to DRX and spheroidization of coarse lamellae \[25–27\]. Moreover, the average grain size decreases with the increase of strain rate. It can be attributed that DRX grains have grown up at the low strain rate of 0.01 s\(^{-1}\).

The inverse pole figures and grain size distributions at a strain rate of 1 s\(^{-1}\) and different temperatures, are shown in figures 5(c1-e1) and (c2-e2), respectively. It can be seen that there are also some DRX grains, as shown by the white arrow. In addition, the average grain size increases with the temperature increasing. Moreover, the temperature has more significant effect on grain size than the strain rate. The main reasons are as follows, high deformation temperature has a positive effect on the formation of \(\beta\) phase, which has more slip systems and is more prone to DRV, so as to weaken DRX. In addition, the deformation temperature has a significant effect on \(\alpha\) phase. The content of lamellar secondary \(\alpha\) phase \((\alpha_s)\) increases with the increasing deformation temperature. In addition, the shape of \(\alpha\) phase change simultaneously. The grains are basically equiaxed when the temperature is

### Table 2. Values of calculated \(n\) and \(Q\) under different deformation conditions.

| Deformation condition | Temperature/°C | \(n\) | \(Q\left(\text{kJ mol}^{-1}\right)\) |
|-----------------------|----------------|------|-------------------------------|
| 0.01–1                | 900–980        | 4.68 | 605.85                        |
|                       | 1020–1100      | 4.87 | 132.44                        |

Figure 3. Relations between: (a) \(\ln \dot{\varepsilon} - \ln \sigma\), (b) \(\ln \dot{\varepsilon} - \sigma\), (c) \(\ln \dot{\varepsilon} - \ln[\sinh(\alpha \sigma)]\), (d) \(\ln[\sinh(\alpha \sigma)] - 1000/T\).

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As the deformation temperature rises to 980 °C, the α_s phase agglomerates and precipitates locally as short needles, as shown by the white circle. When the deformation temperature increased to 1020 °C, higher than the β-transus temperature (998 °C), the β phase transformation behavior is intensified. There are some grain boundary α phase (α_{GB}) grains almost perpendicular to the compression direction, as shown by the black arrow. Besides, some α_s phase grains show a colony, where the lamellar α_s phase is distributed regularly in parallel, and grain orientation is also relatively similar, as shown by the black circle.

3.3.3. Grain boundary structure evolution

Figure 6 shows the grain boundary images and misorientation angle distributions of Ti–6Al–3Nb–2Zr–1Mo alloy at various deformation temperatures and strain rates. Figures 6 (a1–c1) and (a2–c2) show the results of 900 °C at three strain rates. The distribution of misorientation angle shows a significantly high concentration of low misorientation angle due to the aggregation of dislocations during deformation. In addition, the percentage of HAGB is the largest at the strain rate of 0.01 s^{-1} (figure 6(a2)). The low strain rate (0.01 s^{-1}) provides sufficient time for dislocation slip, climb or cross slip, which promotes the DRX and gives rise to a maximum of the percent of HAGBs. With the increase of strain rate (from 0.01 s^{-1} to 0.1 s^{-1}), the dislocation movement is
insufficient, leading to the decrease of DRX behavior and a low transition from LAGBs to HAGBs. Thus, the percentage of HAGBs decreases from 34.92% to 29.64% (figure 6(b2)). However, when the strain rate further increases to 1 s$^{-1}$, the percentage of HAGBs increases slightly from 29.64% to 31.99% (figure 6(c2)). At high strain rate of 1 s$^{-1}$, there is greater dislocation density around the subgrain boundaries, which provide a strong driving force for the shear spheroidization of lamellar grains and nucleation of DRX grain [28], thereby promoting the transition from LAGBs to HAGBs.

The grain boundary maps and misorientation angle distributions at a strain rate of 1 s$^{-1}$ and different temperatures, are shown in figures 6(c1-e1) and (c2-e2), respectively. There is also a portion of LAGBs, which is relevant to the stored dislocation. In addition, the percentage of HAGBs increase prominently from 31.99% to 56.84% as the temperature increases from 900 °C to 980 °C, then increase slightly from 56.84% to 62.52% as the temperature further increased to 1020 °C. In general, high deformation temperature enhances atomic activity and promotes dislocation movement, which consumes a large quantity of dislocations and makes LAGBs decrease. This is the first reason for the increase in the percentage of HAGBs. Besides, as is shown in figures 6(d2) and (e2), the HAGBs show a considerable concentration around 10°, 60° and 90°. According to the BOR of β→α transformation, five misorientation angles among α laths within one prior β grain can be formed theoretically if no variants selection occur [29]. They are 10°/(0001), 60°/(11–20), 60.83°/<1.377, −1.2, 377, 0.359>, 63.26°/<−10, 5, 5, −3> and 90°/<1, −2.38, 1.38, 0> with the ratio of 1:2:3:2:2 [30]. This is the second reason for the increase in the percentage of HAGBs.
4. Discussion

4.1. Activation energy

Activation energy is a significant parameter in the thermal deformation process of metallic materials and is closely related to the deformation behavior [12, 13]. The values of n and Q under various deformation process parameters are shown in table 2. It is clear that the activation energy of Ti–6Al–3Nb–2Zr–1Mo alloy at various deformation temperatures are distinct.

In α+β two-phase region region (860 °C–980 °C), the calculated value of activation energy is 605.85 kJ mol$^{-1}$, which is significantly higher than the activation energy of self-diffusion in α-titanium (204 kJ mol$^{-1}$) [31] and β-titanium (153 kJ mol$^{-1}$) [32] and the diffusion activation energy of Al (92–107 kJ mol$^{-1}$) [33]. As mentioned above, there are a great number of DRX grains around the deformed grain boundary (figure 4). The nucleation of these new grains requires a lot of energy consumption. The DRX behavior should be a significant reason for the substantial increase of Q during thermal deformation. This is consistent with the results of Li. et al [17, 34]. In addition to the DRX behavior, many other explanations have been proposed, including dynamic phase transformation [12], stress concentration [35], and spheroidization of lamellar α phase grains [36].

In single-β region (1020 °C–1100 °C), the calculated value of activation energy is 132.44 kJ mol$^{-1}$, which is close to the self-diffusion energy of β-titanium (153 kJ mol$^{-1}$) [32]. The content of the β phase is considerably large in the deformation temperature range of 1020 to 1100 °C. The β phase is a typical BCC structure with 12 slip systems, which significantly decreases the deformation activation energy. In addition, with the increase of temperature, it is easier for dislocations to slip, climb and cross slip due to the enhancement of atomic diffusion. The above factors are the main reasons for the decrease of activation energy.

4.2. Dynamic recrystallization mechanism

DRX is an essential softening mechanism of titanium alloy during hot deformation, which has a significant effect on the formability of titanium alloy [37]. DRX mechanism can be divided into continuous dynamic recrystallization (CDRX) and discontinuous dynamic recrystallization (DDRX) [38, 39]. A brief description of CDRX is as follows. At the early stage of deformation, a great number of subgrain boundaries form. Then, with the increasing deformation, the orientation angle of the subgrain boundary increases until it reaches the critical value of HAGBs, which causes the transformation of subgrains into new DRX grains. The schematic diagram of CDRX is shown in figure 7(a). Different from CDRX, when the dislocation density or deformation energy storage reaches the critical value, the nucleation of DDRX appears along the deformed grain boundaries [39, 40], and subsequently the DDRX crystal nucleus grows gradually with grain boundary driven by dislocation density. The schematic diagram of DDRX is shown in figure 7(b).
Figure 8 shows the EBSD maps of R1, R2 and R3 regions selected in figures 5 (a1), (b1) and (c1) at 900 °C. It is well known that DRX mechanism is closely related to the grain orientation. It can be seen that new grains have similar orientation. The high point-to-origin cumulative misorientation along the AB, CD and EF directions indicates that the dislocation activity in the grain is considerably increased. In addition, the obvious fluctuation of point-to-point cumulative misorientation is caused by progressive rotation between adjacent subgrains [41]. These microstructure characteristics are consistent with the feature of CDRX [42, 43], that is, the subgrain boundaries are transformed into HAGBs by gradually increasing their misorientation angle. CDRX often causes the increase of the fraction of 10°–15° misorientations (MAGBs), which is regarded as the intermediate stage from LAGBs to HAGBs [44]. Figure 9 shows the frequency of misorientation ranges for Ti–6Al–3Nb–2Zr–1Mo alloy deformed under various deformation conditions. It can be seen that the fraction of MAGBs is only approximately 5% (figure 9(a)). The above phenomena show that CDRX is the DRX mechanism of the alloy deformed at 900 °C, but it is not the dominant DRX mechanism.

It can be observed that there are a great number of new grains with HAGBs along the deformed coarse grain, called necklace structure, which is a significant feature of DDRX [24], as shown in figures 6 (a1), (b1) and (c1). In addition, the orientation between ‘new’ DRX grains (smaller than 1.5 μm) and deformed massive grains (larger than 6 μm) are revealed in the [0001] pole figures, as is shown in figure 10. There is some migration and random grain orientation distribution between deformed grains and new small grains at 900 °C, indicating that DDRX is the dominant DRX mechanism of Ti–6Al–3Nb–2Zr–1Mo alloy at 900 °C. It can be concluded the DRX mechanism is controlled by both CDRX and DDRX at 900 °C, and the latter is dominant.

With the increase of deformation temperature, it is obvious that the fraction of MAGBs decreases from 5.1% to 1.9%, as shown in figure 9(b). The above phenomena indicates that CDRX is weakening gradually. The reason is that the rapid elimination of dislocations by DRV at a higher deformation temperature. As a result, there are not enough dislocations to promote the CDRX process [42]. In addition, the high deformation temperature will accelerate the removal of HAGBs, which can promote the nucleation of the DDRX grain. It can be concluded that DRX mechanism gradually changed to be controlled by DDRX with the increase of temperature to 980 °C and 1020 °C.

Figure 7. Schematic diagrams of (a) CDRX and (b) DDRX mechanism (Different colors represent different grain orientations).
Figure 8. The EBSD maps of R1, R2 and R3 regions: (a1) – (c1) IPF maps, (a2) – (c2) the misorientation profiles along the AB, CD and EF directions.

Figure 9. Frequency of misorientation ranges for Ti–6Al–3Nb–2Zr–1Mo alloy deformed under different deformation condition: (a) 900 °C, (b) 1s^{-1}.

Figure 10. The \{0001\} pole figures of the Ti–6Al–3Nb–2Zr–1Mo alloy at 900 °C and different strain rate: (a) 0.01s^{-1}, (b) 0.1s^{-1}, (c) 1s^{-1} (small grains (less than 1.5 μm in diameter, plotted by color intensity) and coarse deformed grains (large than 6 μm in diameter, plotted by contour line)).
4.3. Texture evolution

Figure 11 shows the ODF of the Ti–6Al–3Nb–2Zr–1Mo alloy at $\phi_1 = 0^\circ$ and $30^\circ$, respectively. The ODF is based on Euler space with a Burger system including three Euler angles of $\phi_1$, $\Phi$ and $\phi_2$ [45], and is effective to describe texture evolution. The texture component can be defined by Miller index, transformed from the Euler angle according to the following formulas [46]:

$$
\begin{bmatrix}
H \\
K \\
I \\
L
\end{bmatrix} = 
\begin{bmatrix}
\sqrt{3}/2 & -1/2 & 0 \\
0 & 1 & 0 \\
-\sqrt{3}/2 & -1/2 & 0 \\
0 & 0 & c/a
\end{bmatrix}
\begin{bmatrix}
\sin \Phi \sin \phi_2 \\
\sin \Phi \cos \phi_2 \\
\cos \Phi
\end{bmatrix}
$$

$$
\begin{bmatrix}
U \\
V \\
T \\
W
\end{bmatrix} = 
\begin{bmatrix}
1/\sqrt{3} & -1/\sqrt{3} & 0 \\
0 & 2/3 & 0 \\
-1/\sqrt{3} & 1/3 & 0 \\
0 & 0 & a/c
\end{bmatrix}
\begin{bmatrix}
-\cos \Phi \sin \phi_1 \sin \phi_2 + \cos \phi_1 \cos \phi_2 \\
-\cos \Phi \cos \phi_2 \sin \phi_1 - \cos \phi_1 \sin \phi_2 \\
\sin \Phi \sin \phi_1
\end{bmatrix}
$$

where $H$, $K$, $I$, $L$, $U$, $V$, $T$, $W$ are Miller index, and $\phi_1$, $\Phi$ and $\phi_2$ are Euler angles. The value of $c/a$ is 1.587 [47], slightly smaller than the values of ideal close-packed hexagonal lattice $c/a$ value (1.633).

Figure 12(a) shows the initial texture of Ti–6Al–3Nb–2Zr–1Mo alloy. It can be known that there are three different texture components in the initial microstructure: $\{1\bar{2}1\ 2\}$$\langle 2\bar{1}1 \ 4 \rangle$ indicated by peak A, $\{1\bar{2}1 \ 1\ 1\}$ (11 07) indicated by peak B and $\{1\ 0\ 1\ 0\}$$\langle 2\bar{1}1 \ 6 \rangle$ indicated by peak C. The Euler angles of the peaks A, B, C are (50°, 63°, 0°), (75°, 80°, 0°) and (54°, 90°, 30°), respectively. The texture intensity of $\{1\bar{2}1\ 2\}$$\langle 2\bar{1}1 \ 4 \rangle$ is strongest about 13.8. It can be concluded that the initial texture is relatively strong and the orientation distribution of $\alpha$ phase grains is not random.

Figures 12(b)–(d) shows the texture evolution expressed as ODF at the deformation 900 °C. At 900 °C, the maximum texture densities are 3.26, 3.94 and 3.69 at the strain rate of 0.01, 0.1 and 1s$^{-1}$, respectively. Texture density is decreased compared with the initial texture, which indicates that the grain orientation of Ti–6Al–3Nb–2Zr–1Mo alloy at 900 °C becomes random. As mentioned above, the orientation of the DRX grains formed by DDRX always different from deformed grains [48], which is the principal reason for the decrease in texture density. In addition, the initial textures are gradually broken, and different types of texture are formed under various deformation conditions. The type of texture gradually evolves to the (0001)//Z0 fiber texture at the deformation temperature of 900 °C and low (0.01 s$^{-1}$) strain rates, as shown in figure 12(b). With the increase of strain rate to 0.1s$^{-1}$, the (0001)//Z0 texture is still significantly strong, and (112 0)//Y0 texture gradually formed in the microstructure. In the process of hot compression, as the increasing deformation, the alloy would flow along X0, Y0 and Z0 under the action of external force, which leads to the start-up of the base and prismatic slip systems along X0, Y0 and Z0. The above phenomenon is the main reason for the formation of the above textures. With the further increase of strain rate, the texture type is different from that at low (0.01 s$^{-1}$) and medium (0.1s$^{-1}$) strain rates. When the strain rate is 1s$^{-1}$, there are mainly $\{\bar{2}113\}$//Y0 and (11 2 0)//Y0 textures, and the former takes the dominant position. At the strain rate of 1s$^{-1}$, it is difficult to satisfy the plastic deformation along the axial direction of high strain rate only by the conventional base and prismatic plane slip. So, slip system along the direction of $\{\bar{2}113\}$ is activated, and quickly dominates to meet the requirements of high strain rate plastic deformation, resulting in the formation of $\{\bar{2}113\}$//Y0 texture.
The texture evolution at the strain rate of 1 s⁻¹ and deformation temperature of 900 °C, 980 °C and 1020 °C, is shown in figures 12(d)–(f). After deformation in the (near) β phase region, texture components are mainly composed of (112 0)//Y0 and (20 2 3)//Y0 textures, with densities of 10.97 and 26.2, respectively. With the increase of deformation temperature, the texture density increases, and reaches the maximum of 26.2 at 1020 °C. High temperature promotes DRV and weakens DRX, resulting in increased texture density. Besides, it is well known that β → α phase transition occurs during the cooling process, and β phase always keeps specific orientation relationship with α phase. This special directional relationship is called the Burgers relationship and expressed as: \{0001\}α//[111]β, (112 0)α//(111)β [49]. The inverse pole figure from Y0 direction deformed at 1020 °C and 1 s⁻¹, is shown in figure 13. Most of the green grains representing the texture of (112 0)//Y0, precipitate in parallel as lath shape, as shown by the white arrow. The β → α phase transformation is a significant reason for the substantial increase in texture density of (112 0)//Y0. In addition, a great quantity of the pink grains, which represent the (20 2 3)//Y0 texture, also precipitate in parallel as lath shape, as shown by the black

**Figure 12.** ODF maps of Ti–6Al–3Nb–2Zr–1Mo alloy: (a) initial state, (b) deformed at 900 °C and 0.01 s⁻¹, (c) deformed at 900 °C and 0.1 s⁻¹, (d) deformed at 900 °C and 1 s⁻¹, (e) deformed at 980 °C and 1 s⁻¹, (f) deformed at 1020 °C and 1 s⁻¹.

**Figure 13.** The inverse pole figure from Y0 direction of Ti–6Al–3Nb–2Zr–1Mo alloy deformed at 1020 °C and 1 s⁻¹.
arrow. Similarly, the formation of the (20 2 3)//Y0 texture is also related to $\beta \rightarrow \alpha$ phase transformation. For HCP structural metals, a 30° rotation around (0001) was reported as the DRX texture components [50], and (20 2 3) can be obtained by (0001) rotate about 40° to (011 0). Therefore, the formation of (20 2 3)//Y0 texture may result from the coordination of DRX grains under the action of the slip system. This conclusion is consistent with the study of Zheng et al [51].

5. Conclusion

Hot compression tests of Ti–6Al–3Nb–2Zr–1Mo alloy were carried out in the temperature range of 900 °C–1100 °C and strain rate range of 0.01–1 s$^{-1}$. The microstructure and texture evolution were investigated by using EBSD technique. The main conclusions are as follows:

1. The flow stress of Ti–6Al–3Nb–2Zr–1Mo alloy is sensitive to the deformation temperature and strain rate during hot deformation. The flow stress decreases with the increasing deformation temperature and decreasing strain rate.

2. The deformation activation energy Q was determined based on the true stress—true strain curves. The value of Q is 605.85 kJ mol$^{-1}$ and 132.44 kJ mol$^{-1}$ in the $\alpha$+ $\beta$ two-phase region and single- $\beta$ region, respectively.

3. At the temperature of 900 °C, the CDRX and DDRX mechanisms are the two DRX mechanisms, and the latter is dominant. With the increase of temperature to 980 and 1020 °C, CDRX gradually weakens, and DRX mechanism changes to be controlled by DDRX.

4. The initial texture is gradually broken, and different textures are formed during hot deformation. The texture evolution was characterized by ODF. The maximum density of texture also changes dramatically. DDRX decreases the texture density of the alloy deformed at 900 °C, and the parallel precipitation of $\alpha$ phase laths rapidly increases the texture density of the alloy deformed at 980 and 1020 °C.

5.1. Declaration of competing interest

The authors declare that they have no known competing financial interests or personal relationships that could have appeared to influence the work reported in this paper.

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Data availability statement

All data that support the findings of this study are included within the article (and any supplementary files).

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