Original Article

Tensile deformation behavior of a near-α titanium alloy Ti-6Al-2Zr-1Mo-1V under a wide temperature range

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ARTICLE INFO

Article history:
Received 1 June 2019
Accepted 7 January 2020
Available online xxx

Keywords:
Titanium alloy
Deformation mechanism
Globularization
Dynamic recrystallization
Electron backscatter diffraction

ABSTRACT

Isothermal tensile tests have been performed to study the mechanical evolution of Ti-6Al-2Zr-1Mo-1V titanium alloy under a wide temperature range from −60 °C to 900 °C and a strain rate of 10−3 s−1. Electron backscatter diffraction (EBSD) tests were used to analyze the evolution of microstructure and deformation mechanisms under different temperatures. The results indicate that deformation mechanisms vary with the deformation conditions. At relatively low temperatures, −60 °C, 23 °C, and 400 °C, dislocation slips mechanism dominates the deformation, even though tensile twinning is detected. While globularization of α laths and dynamic recrystallization (DRX) are dominant at high temperatures, 600 °C and 800 °C, resulting in the flow softening. During the deformation under different temperatures, the β phase plays a crucial role in accommodating deformation between α and β phase by migration of grain boundaries and rotation of grain, leading to a dramatically change in texture of β phase.

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1. Introduction

Ti-6Al-2Zr-1Mo-1V, a near-α titanium alloy, has long been the center of the attention owing to the high strength and toughness, low density, good corrosion resistance and excellent high temperature properties [1–8]. Critical to the success of this alloy group is their capability of being thermomechanically processed and heat-treated in the α + β phase field [9]. That is, good formability. Thus, Ti-6Al-2Zr-1Mo-1V alloy is widely used as load-bearing structural parts, as well as the compressor blades and compressor disks of gas turbine engines [6]. As we know that service conditions, especially, temperature, of the materials play a crucial role in performance. Therefore, it is meaningful to understand how service conditions influence the performance of the materials.

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https://doi.org/10.1016/j.jmrt.2020.01.016
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Due to good formability in $\alpha + \beta$ phase field, hot deformation behavior of Ti-6Al-2Zr-1Mo-1V has been investigated in abundance. These researchers focused their attention on a relatively narrow temperature ranges, $700^\circ$–$970^\circ$ [10–12]. It is true that to know deformation features of Ti-6Al-2Zr-1Mo-1V alloy during hot processing is indispensable, deformation and recrystallization occur simultaneously during hot processing which leads to a complicated evolution of the texture. Efforts have been made to understand the deformation and recrystallization during hot deformation. In general, researchers are more interested in softening mechanisms, for instance, dynamic recovery (DRV), dynamic recrystallization (DRX) and globularization of the $\alpha$ laths; meanwhile, evolution of microstructure and thermal activation energy during the hot deformation of near-$\alpha$ titanium alloys under hot deformation are attracting sufficient attention [10,13]. However, people put Ti-6Al-2Zr-1Mo-1V titanium alloy into center of consideration not only due to good formability in $\alpha + \beta$ phase field, but also owing to its extensive applications in different service conditions. Unfortunately, these investigations which were conducted under a narrow temperature range, usually below phase transformation point 30–300$^\circ$ [10–12], do not reveal how Ti-6Al-2Zr-1Mo-1V titanium alloy performs under low and relatively low temperature. Obviously, this does not benefit for the wide use of the Ti-6Al-2Zr-1Mo-1V alloy. Thus, it is necessary to understand the performance of Ti-6Al-2Zr-1Mo-1V titanium alloy under a wide temperature range.

The objectives of the present work are to: (a) investigate the evolution of mechanical properties, and (b) understand the evolution of the deformation mechanisms during tensile deformation of Ti-6Al-2Zr-1Mo-1V alloy under a wide temperature range to benefit the wide applications of Ti-6Al-2Zr-1Mo-1V titanium alloy.

2. Experiments

Material used in present work is commercial titanium alloy and the nominal composition is Ti-6Al-2Zr-1Mo-1V (wt. %). The $\beta$ transition temperature is confirmed to be 955 ± 5$^\circ$. The rod with a diameter of 300 mm was annealed after rotary swaging at 850$^\circ$ for 2.5 h. Rod like tensile specimens were prepared along the axial direction of the rod, with a gauge length of 25 mm and a diameter of 5 mm. Furthermore, surface of the specimen was polished by silicon carbide papers to avoid some unnecessary factors. The tensile tests were carried out at −60$^\circ$, 23$^\circ$, 400$^\circ$, 600$^\circ$, 700$^\circ$, 800$^\circ$ and 900$^\circ$ using Gleebles–2500 thermal-mechanical simulator along the axial direction of the specimens under a strain rate of 10$^{-3}$ s$^{-1}$, and each specimen was deformed until fracture. Prior to deformation, each specimen was heated to the specified temperature and then held for 15 s before the commencement of the deformation to ensure an establishment of homogenous temperature field in the specimen. The temperature was controlled by a thermocouple welded to the central part of the specimen surface. After deformation, all specimens were immediately water quenched in order to maintain the deformed structure.

The areas selected for microstructural observation were adjacent to the necking areas to investigate the uniform deformation characters. For EBSD tests (OXFORD EBSD acquisition camera and the OXFORD-AZTEC online acquisition software and JEM-7001 F instrument equipped with an OIM Analysis system operated at 20KV), the specimens were wet-ground with silicon carbide papers and then mechanically polished with OPs polishing solution. The grain boundaries with misorientation between 1$^\circ$ and 15$^\circ$ are defined as low-angle boundaries (LGBs), and those of misorientation above 15$^\circ$ as high-angle boundaries (HGBs). Inverse pole Figures obtained from the EBSD data are taken alone the tensile direction (RD/tensile direction).

3. Results and discussion

3.1. Mechanical behavior

Fig. 1 shows the typical true stress-strain curves of Ti-6Al-2Zr-1Mo-1V titanium alloy under a strain rate of 10$^{-3}$/s under different temperatures. The stress shows a rapid increase till the stress reach the yield point during the initial deformation stages under all temperatures. However, the yield stress of these curves become lower with increasing temperature as shown in Fig. 1, indicating a sharp sensitivity to the temperature. At high temperature (800$^\circ$ and 900$^\circ$), the initial deformation stages are followed by a decreasing tendency after reaching the peak value (yield point). Differently, the true stress-strain curve at 800$^\circ$ shows a rapidly decreasing tendency, while a steadily decreasing tendency occurs under 900$^\circ$. However, distinct strain-hardening phenomena occurs under relatively low temperature (below 400$^\circ$). Furthermore, the flow stress keeps roughly constant at 600$^\circ$. For relatively low temperature, the strain-hardening rate curves exhibit mild change during the plastic deformation, and, meanwhile, the curves have a similar positive strain-hardening behaviors as shown in the inset image in Fig. 1, which means that they may share similar deformation mechanisms at relatively low temperature [14–17], we will discuss this topic in hind part. In contrast, under high temperature, true stress-strain curves express the apparent flow-softening phenomena, and naturally, the strain-hardening curves show negative strain-
3.2. Microstructural evolution

3.2.1. Initial microstructure

The initial microstructure and texture detected by EBSD tests of as-received Ti-6Al-2Zr-1Mo-1V titanium alloy are shown in Fig. 2. In Fig. 2(a), black areas indicate the β phase, two morphologies of α phase can be seen, equiaxial and layered α phase. Fig. 2(b) shows the grain boundary map of Ti-6Al-2Zr-1Mo-1V alloy. Blue and red lines indicate the HABs and LABs, respectively. It can be seen that even though as-received Ti-6Al-2Zr-1Mo-1V alloy was annealed at 850° for 2.5 h, however, LABs can be detected. This illustrates that severe deformation caused by the rotary swaging is not totally restored. Fig. 2(c) and (d) show the inverse pole figures (IPFs) of α and β phase, respectively. X, Y, and Z represent the transverse direction, rolling direction, and normal direction, respectively. Although textures exist in initial Ti-6Al-2Zr-1Mo-1V alloy, peak values of textures in α and β phase are 1.77 and 2.79, which indicates that textures existing in initial Ti-6Al-2Zr-1Mo-1V alloy is too weak to consider.

Fig. 3 shows the further features of the initial Ti-6Al-2Zr-1Mo-1V alloy. Fig. 3(a) and (b) show the Schmid factor (SF) profiles of possible slip systems in α and β phases, respectively. For α phase, only pyramidal and 1st pyramidal slip system are preferred. However, SF values of the possible slip systems of β phase are above 0.3, which indicates that all slip system can be possibly active [18]. The misorientation angle profile [Fig. 3(c)] indicates that HABs are predominant, even though LABs can be detected, consisting with the grain boundary map in Fig. 2(b). Grain size distribution in Fig. 3(d) illustrates that size distribution of α phase is well fitted by normal distribution and the grain size is confirmed to be ~15 μm. Fig. 3(e) and (f) display the recrystallization distribution histogram of α and β phases of initial Ti-6Al-2Zr-1Mo-1V alloy, respectively. For α phase, recrystallized and substructured α phases are dominant, only a little deformed α phase occurs, indicating that deformed α phase has been nearly totally transformed during annealing. However, the frequency of deformed β is much higher than deformed α, and substructured β phase sharply decreases. This indicates that the extent of recrystallization of β phase is much lower than that of α phase.

3.2.2. Deformed microstructure under different temperatures

Fig. 4 show the EBSD images of the deformation microstructures of Ti-6Al-2Zr-1Mo-1V titanium alloy under different temperatures. At relatively low temperatures (below 400°), tensile twinning with a typical misorientation angle of 86.3°
was detected as shown in Fig. 4(a)–(c). Tensile twinning is a usual twinning type in HCP alloys, such as magnesium alloys [19,20], pure titanium [21], and Ti-6Al-2Zr-1Mo-1V alloy [22]. However, frequency of twinning is not that high when Ti-6Al-2Zr-1Mo-1V alloy was deformed under −60°, 23°, and 400°, unlike the situation in pure Ti or Mg alloys. This may demonstrate that twinning is not a prevalent deformation mechanism when deformed under relatively low temperatures in current alloy. With the increasing temperature, the β phase (black areas) becomes increasing after the hot deformation is finished as shown in Fig. 4(d) and (e) and this phenomenon is also found in other investigations when hot deformation was performed in near-α titanium alloys [10]. But what differs from the situation in relatively low temperatures is that α laths become fragmentized under high temperatures. As for the equiaxed α grains, the grain size does not apparently change, compared to initial state [see Fig. 2(a)]. Fig. 5 displays the evolution of α grain size when deformed under different temperatures. From Fig. 5(a)–(c), the average α grain size under relatively low-temperatures deformation can be confirmed to be 13 μm, 14 μm, and 14.5 μm, indicating a small deviation from the initial state. However, under high temperature, the grain size of α phase declines sharply as shown in Fig. 5(d) and (e).

Fig. 6 shows the inverse pole figures (IPFs) of deformed microstructure under different temperatures. It can be seen that the maximum texture intensity of the α phase decreases monotonously from 4.396 to 1.862 at relatively low temperatures. With further increasing temperature, the maximum texture intensity of the β phase increases to 3.06 and then decreases to 2.47. The maximum texture intensity expresses complicated change at a wide temperature range from −60° to 800°, indicating different modes of microstructural evolution. The texture peak can be observed around the <10 1 0> [see Fig. 6(a)], with the increasing temperature, texture peak rotate to the direction between <0001> and <10−10> as shown in Fig. 6(b), and some grains rotated towards the position between <0001> and <2−1 1 0> directions as shown in Fig. 6(c),
which leads to the weakening and relatively randomization texture after deformation. At high temperature, the texture becomes increasingly randomized [see Fig. 6(d) and (e)], and no obvious texture can be detected even though the maximum texture intensities are higher than that under relatively low temperatures (400 ◦C), which indicates that the deformation mechanisms may change under high temperatures compared to relatively low temperature.

3.3. Deformation mechanisms

3.3.1. Under relatively low temperatures

At the relatively low temperatures, true stress–strain curves express the evident strain-hardening behaviors and can be well fitted by straight lines as indicated in Fig. 1. Even though deformed at different temperatures under relatively low temperatures, however, the slopes of the fitted lines illustrate a similar values. This may indicate that they share a common deformation mechanisms during deformation at relatively low temperatures [14–17]. Tensile twinning was detected during the deformation under relatively low temperatures as shown in Fig. 4(a)–(c), however, note that the content of tensile twinning is not high and twinning bands just emerge in individual grain. Fig. 7 shows the high-magnification images of the tensile twinning detected at −60 ◦C and 400 ◦C. The misorientation profiles definitely demonstrate that twinning bands in Fig. 7(a) and (d) are typically tensile twinning. However, the Kernel average misorientation (KAM) maps [Fig. 7(b) and (e)] show that color in the areas adjacent to the twinning bands are mild, which indicates that dislocations are not severely piled up at the twinning boundaries. This phenomenon is in conflict with situation that twinning bands are the main reason to strain-hardening behavior [23,24]. Furthermore, plastic deformation under relatively low temperatures possesses a roughly constant strain-hardening rate as shown in Fig. 1. For the twinning mechanism, the flow stress usually shows a complicated variation tendency with strain, and its outline is somewhat like an “S” [25–27]. More importantly, the strain-hardening rate also expresses a complex tendency, usually its outline is somewhat like a mirror symmetric “N”

Fig. 4 – EBSD images of the deformed microstructures under different temperatures. (a)–(c) Under relatively low temperatures, −60 ◦C, 23 ◦C, and 400 ◦C, respectively. (d) and (e) Under high temperatures, 600 ◦C and 800 ◦C, respectively. Black areas in the EBSD images correspond to the β phase.
Fig. 5 – Grain size distribution after deformation at different temperatures. (a)–(c) Under relatively low temperatures, −60 °C, 23 °C, and 400 °C. (d) and (e) Under high temperatures, 600 °C and 800 °C.

Fig. 6 – Inverse pole figures (IPFs) taken along the rolling direction (Y direction: Loading Direction) of the deformed structures under different temperatures. (a)–(c) α phase: under relatively low temperatures, −60 °C, 23 °C, and 400 °C, respectively. (d) and (e) β phase: under high temperatures, 600 °C and 800 °C, respectively.

[28,29] Of course, these features are not only can be found in the metal deformed via twinning, but also in the metal deformed via stress-induced transformation, such as stress-induced martensite in metastable β titanium alloys [30]. When it comes to the dislocation slips mechanisms, the plastic deformation usually show a steady tendency with the increasing strain and the strain-hardening rate curve possesses a plateau [31,32]. When deformed under relatively low temperatures, the situation in current work has a definite deviation from twinning mechanism, but it is similar with the fact under dislocation slips mechanisms.

To further clarify the deformation mechanism at relatively low temperatures, three different areas selected in the specimens deformed at −60 °C, 23 °C and 400 °C are shown in Fig. 8.
(IQ maps and KAM maps). Black areas in the IQ maps indicate the β phase has a low value of quality, indicating a more severe lattice distortion [33]. Red areas in the KAM maps represent a higher misorientation angle, indicating a higher density of geometrically necessary dislocations (GNDs) [33,34]. Obviously, the red areas are consistent with black areas (β phase) in the IQ maps in Fig. 8, which illustrates a more severe deformation in β phase as well. It is not difficult to understand the reason that β phase possesses more dramatic plastic deformation. Just similar with the situation in duplex steels, the deformation concentrated in the “softer” zone, austenite [34]. This is mainly attributed to the fact that the austenite accumulates more deformation than the “harder” phase, where the martensite is much stronger than the austenite, so that the strain concentrates in the austenite. But for the α phase, deformation usually concentrated in grain boundaries as shown in Fig. 8, and these GNDs accumulated in the α grain boundaries to accommodate deformation and made plastic deformation in α and β phase compatible during the deformation. The average misorientation angle θ in three different areas was calculated to be roughly 4° for all α phases, while 2° at −60° and 400°, ~2.25° at 23°, for β phase, from the KAM profiles shown in Fig. 8(d) and (e). Based on the strain gradient model by Gao et al. [35], they define a GND array for simple cylinder torsion. Assuming a series of twist subgrain boundaries in the cylinder, each containing two perpendicular arrays of screw dislocations, the misorientation angle θ is related to the GND density $\rho_{\text{GND}}$, $\rho_{\text{GND}} = 2\pi u b$, where $u$ is the unit length and $b$ is the magnitude of the Burgers vector. Finally, the density of GNDs is confirmed to be about $2 \times 10^{15}/\text{m}^2$ in α phase, and $1 \times 10^{15}/\text{m}^2$ at −60° and 400°, $1.125 \times 10^{15}/\text{m}^2$ at 23°, in β phase. We ignore the statistical storage dislocations (SSDs) when we calculate the density of dislocations. Even though SSDs are also generated when deformed, however, it is reported that the density of GNDs dominated the total dislocation density in the small and medium plastic deformation of the polycrystalline metals and alloys [36]. The present maximum amount of deformation (below 30%), at relatively low temperatures, is within the medium plastic deformation regime. At relatively low temperatures, the densities calculated of dislocation are similar with each other, illustrating that they have a similar deformation feature (similar strain-hardening behavior), that is, dislocation slip mechanism. These results indicate that the predominant deformation mechanism at relatively low temperatures is the slip mechanism rather than twinning mechanism, even though tensile twinning was detected.

Fig. 8(f) shows the (0001)α and (110)β pole figures. As we know, α and β phase obey the Burgers orientation relationship (BOR), {110}β//[0001]α, and <112>β//<11−20>α [37]. However, when deformed at 23°, the (0001)α and (110)β pole figures indicate that such BOR did not occur. This means that there is no strict BOR between α and β phase, and this phenomenon usually occurs under hot deformation conditions [38]. As for the current situation, it may be caused by the excessive accommodation of deformation between α and β phases. The average misorientation angle in β phase at 23° is higher than others, indicating that β phase was subjected to more severe deformation when accommodating the deformation. To accommodate the deformation between two phases, β phase deformed dramatically, and made a deviation from the strict BOR between α and β phase. In addition, the Schmid factor of the deformed β phase under relatively low temperatures is shown in Fig. 9. It can be seen that values of the SF distribute mainly in the range from 0.4 to 0.5. Grain boundaries migration and grain rotation...
is possible under high SF value, and meanwhile, make them a primary deformation mechanism [39,40]. From the (100) β pole figure shown in Fig. 9(d) and (e), the texture rotated about 90°, from (110)<110> texture, rotated Goss texture, to (110)<100>, Goss texture, as indicated by open black frame and open black circle in (d) and (e), respectively. The change of the texture of
the β phase demonstrate that the β grains rotated dramatically due to high SF and finally made a sharply different texture. In fact, the “softer” phase β phase plays the role of “lubricant” during the deformation. Through migration of the grain boundaries and rotation of the grains, β grains offset the difference in ability to deformation between α and β phase, leading to the existence of unprecise BOR between α and β phase. In addition, high SF of β phase is independent of the texture, even at high temperature, the SF is still high, indicating essence role of the β phase during the deformation.

3.3.2. Under high temperatures
At high temperature, no tensile twinning could be found as shown in Fig. 4(d) and (e), as reported by Williams [41] that twinning is suppressed with increasing temperature. Furthermore, flow stress keeps roughly constant at 600 °C, and flow softening occurs at higher temperatures, 800 and 900 °C, as shown in Fig. 1. As for softening mechanism, many factors may contribute to the flow softening, such as dynamic grain rotation or platelet kinking, heat generation, flow instability, texture and substructure evolution [10]. For the present study, the main softening mechanism is globularization, DRX also plays a role in it, but not dominant. Fig. 10 shows fraction of recrystallization at high temperatures. The areas in blue, yellow and red represent the recrystallized, substructured and deformed grains, respectively. And the colors have the same meaning in the histogram. It is obviously seen that the fraction of recrystallization is higher in initial state than that in deformed at 600 or 800 °C as shown in Fig. 10(a)−(c). This is due to initial rod was subjected to annealing at 850 °C for 2.5 h. A low content of deformed structure indicates that deformed areas have transformed to the recrystallized and substructured grains. However, when deformed at 600 and 800 °C, the contents of deformed structure become sharply higher than that in initial state. In addition, a small increment in recrystallized structure occurs when temperature increased to 800 °C, meanwhile, the contents of deformed and substructured areas slightly changed, as shown in Figs. 10(b)−(e). For the β phase, no evident changes occur during the hot deformation [Fig. 10(g) and (h)], compared to the initial state of β phase as shown in Fig. 10(f). In addition, at high temperatures, the SFs profiles of all three possible slip systems in β phase distributed mainly in the range from 0.4 to 0.5, similar with the situation in Fig. 9(a)−(e), as mentioned in part 3.3.1. Taking the fraction of recrystallization and high SFs in the β phase into account, β phase plays a crucial role in accommodating the deformation between the α and β phase, the high SFs make grain boundaries migration and grain rotation possible, similar with the situation at relatively low temperatures, indicating an essential effect of β phase during the deformation under high temperature range. Fig. 11 shows the misorientation angle profiles of α phase deformed under high temperatures. It can be seen a similar tendency occurs at −60°, 23°, and 400°, the peak values distribute mainly below 10° and above 85°. However, at 23°, the fraction of the peak value below 10° is lower than those at −60° and 400°. As mentioned above, β phase excessively accommodates the deformation at temperature of 23°, that is, β phase bears more plastic deformation, and correspondingly, α phase undergoes less deformation than those at −60° and 400°, resulting in a relatively low fraction of LABs. However, at high temperatures, misorientation angle main distributed above 20°, high angle. It is widely accepted that
Fig. 10 – (a) Recrystallization map of α phase of initial specimen. (b) and (c) Recrystallization maps of α phase deformed under 600 ° and 800 °. White areas correspond to β phase. (d) and (e) α phase: recrystallization histograms of the recrystallized, substructured, and deformed grains at 600 ° and 800 °. (f)–(h) β phase: recrystallization histograms of the initial specimens, deformed structure at 600 °, and deformed structure at 800 °.

DRX will decrease the content of the LABs and make HABs increase [2,10]. However, the fraction of DRX is rather low as shown in Fig. 10(b)–(e), illustrating that DRX may not be the main reason. Also, it has been reported that globularization of α laths plays an important role in hot deformation and makes softening mechanism operated [13,42]. From Fig. 4(d) and (e), it can be obviously seen that α laths become fragmented, especially in Fig. 4(e), and finally make the grain size of α sharply decrease as shown in Fig. 5(d) and (e), compared to initial state [Fig. 3(d)] and deformed at relatively low temperatures [Fig. 5(a)–(c)]. Thus, from the results listed above, it can be concluded that the main softening mechanism at high temperature is proved to be globularization of α laths. Meanwhile, the DRX seems to become increasingly stronger with
increasing temperature as shown in Fig. 10(b) and (c). This is mainly due to the sharp decrement in stress at high temperature, for example, peak value ~50 MPa at 800 °C, the original α lamellae cannot be easily decomposed by forming high angle grain boundaries though the dislocation accumulation, which is the globularization mechanism at 600 and 800 °C. Thus, the DRX began to become stronger to accommodate the hot deformation. But note that main deformation mechanism under high temperatures is the globularization of α laths as proved above, even though DRX becomes stronger.

When it comes to the apparent activation energy of deformation, it can also be utilized to elaborate the trait of deformation. The relationships between temperature, strain rate, flow stress, and activation energy can be described by the Arrhenius equation [43]:

\[ \dot{\varepsilon} = A \sigma^n \exp\left(-\frac{Q}{RT}\right) \]  

(1)

where \( R \) is the universal gas constant (8.314 J mol\(^{-1}\) K\(^{-1}\)), \( T \) is the absolute temperature (K), \( A \) is the material constant, \( n = 1/m \) is the stress index, \( m \) the strain-rate sensitivity index, \( Q \) the activation energy of deformation (KJ/mol). When strain rate is constant, by taking natural logarithm from both sides of Eq. (1), the following expression can be written as:

\[ Q = nR \frac{\partial \ln \dot{\varepsilon}}{\partial (1/T)} \]  

(2)

when \( \dot{\varepsilon} \) was given as the value of stress at yield point, \( Q \) can be deduced from the slope in a plot of \( \ln \sigma \) as a function of \( 1/T \), as shown in Fig. 12. It is obviously seen that the plot in Fig. 12 shows a distinct deviation from those obtained from conventional hot deformation (deformed in a narrow temperature range). Usually, a linear fitted line can represent the tendency of the plot, however, in present work, the plot shows a two-stage linear tendency. The slope under high temperatures deformation (600, 700, 800, and 900 °C) is definitely higher than that under relatively low temperatures (~60, 23, and 400 °C). And such high value does not present any true physical significance but was rather an artifact of the fitting of data for two-phase materials [44]. It has been reported that DRX including continuously dynamic recrystallization (CDRX) and discontinuously dynamic recrystallization (DDRX), phase transformation, the globularization of the initial lamellar microstructure and starting microstructure with lamellar shape [10] can increase the activation energy. As for the situation in current study, high slope of the fitted line at high temperatures (indicated by blue) is caused by the globularization of the initial lamellar microstructure and increasingly
Fig. 13 – Schematic diagram of the deformation mechanisms evolution at different temperatures in Ti-6Al-2Zr-1Mo-1V titanium alloy.

Stronger DRX. At low temperature, slip mechanism dominates the deformation as proved in the part 3.3.1, thus, a well linear tendency can be obtained under such temperature range as shown in Fig. 12 (indicated by red). Interestingly, the value obtained at 600°C simultaneously satisfies linear relationships at high and low temperature range as shown in Fig. 13. The possible reason is that the DRX becomes stronger above 600°C, leading to a significant effect on activation energy. Even though we can obtain an approximate value of activation energy at low temperature range by fitting the splashes, in fact, the value we obtained is still higher due to the starting microstructure with lamellar shape. When we take the value of $n$, stress index, to be 4.5 from the research similar with present study [10], the $Q$ can be confirmed to be ~440 kJ/mol, which is higher than the activation energy values for self-diffusion of α-Ti (204 kJ/mol) [45] and β-Ti (153 kJ/mol) [46]. This indicates that starting microstructure with lamellar shape dose have a strong impact on activation energy and makes it higher.

3.3.3. Activation sequence of deformation mechanisms under different temperatures

The establishment of the activation sequence of the different deformation mechanisms is of great importance for the optimization of the macroscopic mechanical performance of these materials. From the results demonstrated above, we have proved that different deformation mechanisms are activated during deformation with the elevated temperature and a schematic diagram is given to model the mechanism operated during the deformation under a wide temperature range as shown in Fig. 13. At low temperature, dislocation slips is the predominant mechanism, even though tensile twinning is also detected. With increasing temperature, the twinning disappears and main deformation mechanism transforms to be globularization of α laths, meanwhile, DRX is also operated and becomes stronger with the increasing temperature.

4. Conclusion

Tensile deformation behaviors of a near-α titanium alloy, Ti-6Al-2Zr-1Mo-1V, were studied by tensile deformation under a wide temperature range. The deformation mechanisms activated under different temperatures were investigated by EBSD tests. At low temperatures, dislocation slips is the main deformation mechanism, even though tensile twinning is also detected in α grain. At high temperatures, globularization of α laths and DRX occur simultaneously, which plays a crucial role in flow softening behavior. However, globularization of α laths dominates the deformation at high temperatures. In addition, starting microstructure with lamellar shape, DRX, and globularization of α laths have a strong effect on activation energy of deformation, leading to a higher value of activation energy. As for the β phase, it plays a positive role to accommodate the deformation between α and β phases through migration of grain boundaries and grain rotation whether at low or high temperatures.

Author contributions

The co-first author J.F. Xiao analyzed the datas and wrote this paper. The co-first author F. Hao performed the involved experiments. Both of two authors made the equivalent contributions to this paper. The rest of authors gave the positive and constructive suggestions during experiments, and data analysis, and their contributions are also important and indispensable.

Conflicts of interest

The authors have no conflicts of interest to declare.
Acknowledgement

The authors acknowledge the support of the National Natural Science Foundation of China (No. 51671031).

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