



Research progress on microstructure evolution and hot processing maps of high strength β titanium alloys during hot deformation

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Abstract: High strength β titanium alloys are widely used in large load bearing components in the aerospace field. At present, large parts are generally formed by die forging. Different initial microstructures and deformation process parameters will significantly affect the flow behavior. To precisely control the microstructures, researchers have conducted many studies to analyze the microstructure evolution law and deformation mechanism during hot compression. This review focuses on the microstructure evolution of high strength β titanium alloys during hot deformation, including dynamic recrystallization and dynamic recovery in the single-phase region and the dynamic evolution of the α phase in the two-phase region. Furthermore, the optimal hot processing regions, instability regions, and the relationship between the efficiency of power dissipation and the deformation mechanism in the hot processing map are summarized. Finally, the problems and development direction of using hot processing maps to optimize process parameters are also emphasized.

Key words: high strength β titanium alloy; hot deformation; microstructure evolution; hot processing map

1 Introduction

Titanium is an allotropic element and can appear in two crystal structures. Pure titanium is composed of the close-packed hexagonal (HCP) α phase at a temperature lower than 882 °C and the body-centered cubic (BCC) β phase at a temperature higher than 882 °C [1]. In the production of alloys, each element added to pure titanium will determine the mechanical properties by affecting the stability of α and β phases. Among these elements, Al, Ga, O, N and C are α stabilizers [2], V, Mo, Nb, Fe, Cr and

Ni are β stabilizers [3], and Sn and Zr are neutral elements [4].

According to the relationship among the phase composition, β stable elements and content of the annealed state, titanium alloys can be divided into three categories: α titanium alloy, $\alpha+\beta$ titanium alloy and β titanium alloy. β titanium alloy has good toughness and hardenability and has been widely used in large-scale bearing components in the aviation field. The β phase titanium alloy contains a large amount of β phase stabilizing elements. β phase stabilizing elements are divided into two types: the eutectoid elements and isomorphous

elements [5]. The eutectoid elements are Cr, Mn, Fe, Co, Ni and Cu, while the isomorphous elements are Mo, V, Ta and Nb. For a given titanium alloy composition, a known and useful parameter that characterizes the stability of the β phase is the molybdenum equivalent ($[\text{Mo}]_{\text{eq}}$). This quantity is a combined measure of the effects of β phase stabilizing elements, α phase stabilizing elements, and neutral elements contained in titanium alloys on the β phase stability [6]. The equation is given by [5]:

$$[\text{Mo}]_{\text{eq}} = 1.0\text{Mo} + 0.67\text{V} + 0.44\text{W} + 0.28\text{Nb} + 0.22\text{Ta} + 1.6\text{Cr} + 2.9\text{Fe} + 1.54\text{Mn} + 1.25\text{Ni} - \text{Al} \quad (1)$$

Generally, $0 \leq [\text{Mo}]_{\text{eq}} \leq 10$ is near β titanium alloy, $8 \leq [\text{Mo}]_{\text{eq}} \leq 30$ is metastable β titanium alloy, and $[\text{Mo}]_{\text{eq}} \geq 30$ is stable β titanium alloy [7]. The corresponding relationship between the mechanical properties and $[\text{Mo}]_{\text{eq}}$ of some typical high strength β titanium alloys is shown in Fig. 1.

High strength β titanium alloys represented by metastable β titanium alloys and near β titanium alloys have good mechanical properties and are widely used in large load bearing components in the aerospace field [26]. At present, the processing method for the above large parts is mainly die forging. Compared with traditional steel and aluminum alloys, titanium alloys are very sensitive to deformation temperature, strain rate and strain. Therefore, it is of great significance to master and understand the microstructure evolution and deformation mechanism of titanium alloys during

deformation. The interaction of work hardening and dynamic softening during hot deformation of high strength β titanium alloys can control the microstructural evolution and mechanical properties [27] and influence the flow behavior. At the same time, the difference in deformation process parameters and initial microstructure [28] further affects the flow behavior. Only by combining the microstructure can we deeply analyze the influence mechanism of deformation process parameters on the flow behavior and then provide guidance for the forging process. It is well known that the forging methods of high strength β titanium alloys are mainly $\alpha+\beta$ forging and β forging [29]. β forging has very low deformation resistance, good metal fluidity, and low forging cost. In addition, the refinement of β grains is beneficial to weakening the micro-texture and improving the performance of ultrasonic flaw detection [29]. The main purpose of $\alpha+\beta$ forging is to retain the primary α phase and obtain a good match of strength and toughness. The flow behavior of these forging processes is accompanied by complex microstructure evolution. Therefore, it is necessary to understand the microstructure evolution and deformation mechanism during hot deformation. For β titanium alloys, the existing review mainly focuses on the application of phase stability [30], heat treatment [1,31], and aircraft structures [12]. This paper reviewed the microstructure evolution, deformation mechanism, and hot processing map of high strength β titanium alloys during hot deformation.

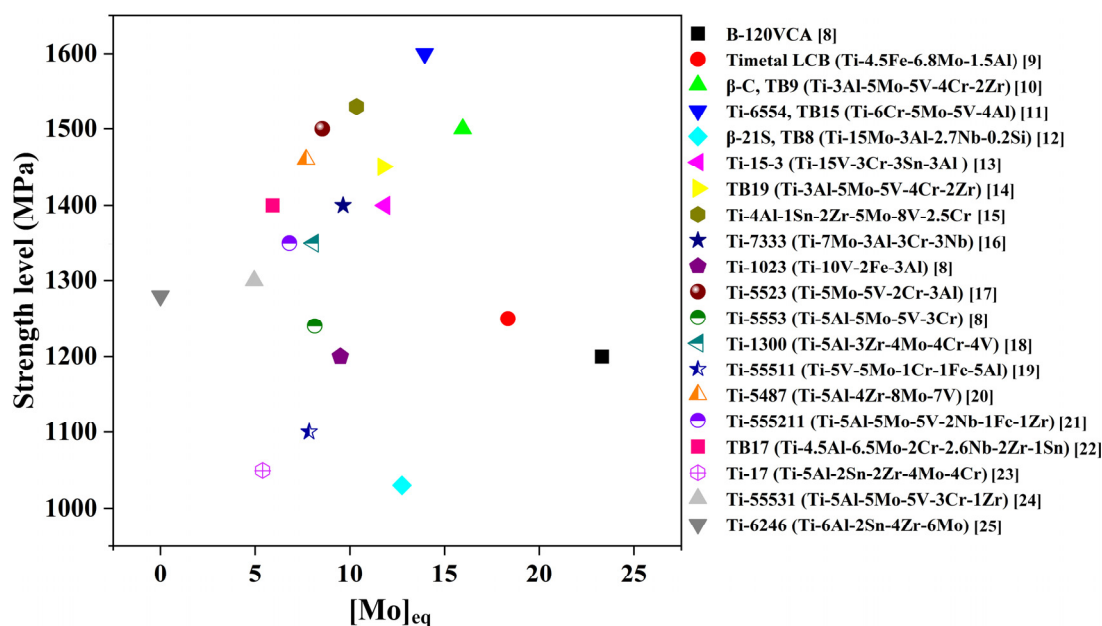


Fig. 1 Relationship between strength level and $[\text{Mo}]_{\text{eq}}$ of different high strength β titanium alloys

Firstly, as the traditional softening mechanism of metal materials, dynamic recovery (DRV) and dynamic recrystallization (DRX) are the main ways to reduce deformation resistance. DRX is divided into continuous dynamic recrystallization (CDRX), discontinuous dynamic recrystallization (DDRX) [32] and geometric dynamic recrystallization (GDRX) [33]. DRV and DDRX often appear in the β phase region. In contrast, the dynamic behavior of the α phase, including dynamic phase transformation, kinking, globularization and precipitation, is a typical characteristic of hot deformation in the two-phase region. The dynamic globularization mechanism of the α phase has been deeply studied as a hot topic in frontier research. The deformation conditions and initial state of the sample significantly affect the above-mentioned softening mechanism. Furthermore, texture evolution has also been mentioned in some studies. Subsequently, the hot processing maps of different high strength β titanium alloys are summarized, including the optimal hot processing regions, the instability regions, the relationship between the efficiency of power dissipation and the deformation mechanism, and some typical instability phenomena. Finally, the development direction and difficulties needing more research efforts in this field are summarized and prospected.

2 Microstructure evolution of β titanium alloy during hot compression

Microstructure evolution is the most common phenomenon in various metal materials during hot deformation. Due to the difference in the initial

microstructure and deformation parameters, β titanium alloys usually exhibit multiple deformation mechanisms during the deformation process. The deformation process parameters significantly affect the microstructure (phase composition, morphology, the distribution and size). Meanwhile, the microstructure evolution will determine the mechanical properties of titanium alloy forgings. Therefore, it is necessary to understand the microstructure evolution law and deformation mechanism during hot deformation to accurately control the microstructure and mechanical properties of titanium alloys. Figure 2 shows the main deformation mechanism of high strength β titanium alloy under different initial microstructures and deformation conditions. DRX and DRV are dominant in the single-phase region, while α phase evolution is dominant in the two-phase region. In this section, different deformation mechanisms are analyzed in detail.

2.1 Dynamic recovery

The dynamic recovery (DRV) is mainly manifested as the cross slip of screw dislocations and climb of edge dislocations, resulting in the annihilation of dislocations. For high stacking fault energy metals, DRV is the only softening mechanism in the hot deformation process. For metals with low stacking fault energy, when the deformation degree cannot reach the critical condition for DRX, only DRV occurs. DRV is a common softening mechanism during the hot deformation of β titanium alloys [29]. The recovery process promotes the rearrangement of dislocations and the generation of subgrains [34]. In many

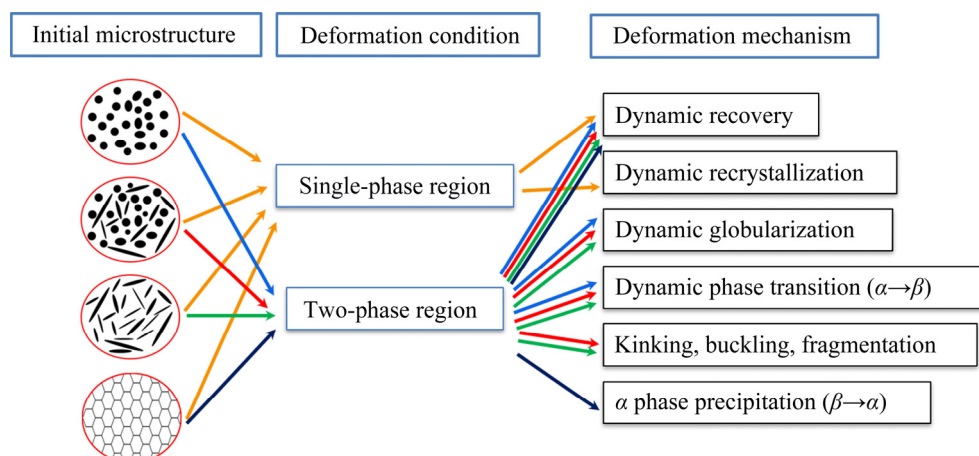


Fig. 2 Main deformation mechanism of high strength β titanium alloy under different initial microstructures and deformation conditions

studies, DRV has the following characteristics during the hot deformation of β titanium alloy [21,29,34–46]:

(1) The apparent activation energy for deformation is close to the self-diffusion activation energy of pure Ti, and the stress–strain curve is a typical recovery curve.

(2) The microstructure is not uniform after deformation. Many elongated grains with zigzag boundaries and a large number of subgrains can be observed.

(3) The dislocation walls are formed by dislocation tangles, accompanied by many low angle grain boundaries (LAGBs).

Work hardening plays an essential role in DRV, and the work hardening rate can be related to DRV. The relationship between the stress (σ) and the work hardening rate in the DRV process (Q_{DRV}) can be expressed as [47–50]:

$$\theta_{\text{DRV}} = \frac{\alpha M G b k_1}{2} - \frac{k_2 \sigma}{2} \quad (2)$$

where α is the material constant, G is the shear modulus, b is the amplitude of Burgers vector (2.86×10^{-10} m), M is the Taylor factor that realizes

the conversion from shear strain γ to normal strain ε , $\varepsilon = M\gamma$, k_1 is a constant related to the material properties ($k_1 = 2\theta_{\text{II}}/(\alpha G b)$), θ_{II} is the slope of the stress–strain curve in the second stage ($\theta_{\text{II}} \approx G/200$), k_2 is the dislocation annihilation rate ($k_2 = k\alpha M G b k_1/\sigma_p$, k is the constant of proportionality, and σ_p is the peak stress). It can be found from the equation that the work hardening rate has a linear relationship with stress.

To reveal the DRV mechanism of β titanium alloys from a microscopic perspective, many researchers have studied the microstructure changes of β titanium alloy during hot deformation using optical microscopy (OM) and electron backscattered diffraction (EBSD). MATSUMOTO et al [43] and MENG et al [51] found that Ti-5553 alloy and Ti-6246 alloy had elongated β grains and a small number of zigzag boundaries in the β phase region and the $\alpha + \beta$ phase region, as shown in Figs. 3(a, b). Although recrystallization characteristics appear, the softening mechanism is dominated by DRV. Because of the high stacking fault energy of the BCC structure, the dislocations easily slip and climb, and significant dislocation pile-up does not easily form. The dislocation tangle transforms into

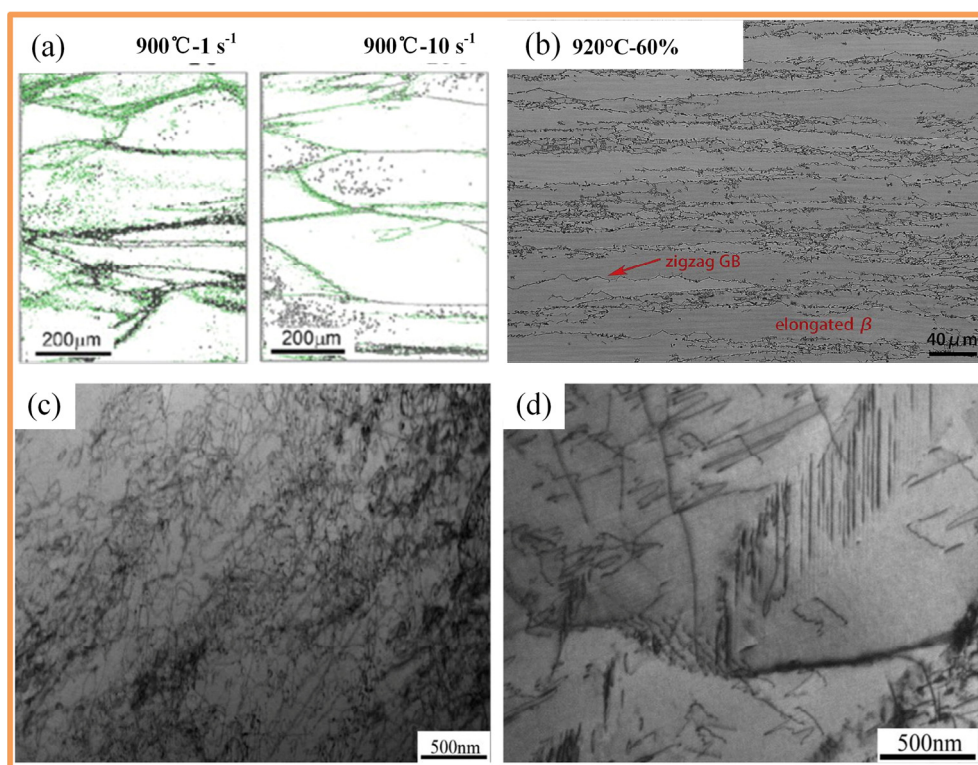


Fig. 3 Typical phenomena of DRV: (a) Elongated β grains in Ti-5553 alloy [43]; (b) Zigzag grain boundaries in two-phase region of Ti-6246 alloy [51]; (c, d) Dislocation evolution of Ti-5553 alloy at different strain rates of 0.1 s^{-1} (c) and 1 s^{-1} (d) [29]

the cellular structure or subgrain, resulting in the appearance of many LAGBs. The proportions of LAGBs in the α and β phases are 71.9% and 87.4% of Ti-52244 alloy deformed at 810 °C and 0.01 s^{-1} , respectively [52]. In Ti-55511 alloy [53] and Ti-55531 alloy [29], dislocation climbing and dislocation entanglement were observed by transmission electron microscope (TEM) characterization. Among them, Fig. 3(c) shows the TEM image in the β phase region of the Ti-55531 alloy at a strain rate of 0.1 s^{-1} , from which an obvious dislocation tangle can be observed. When the strain rate increases to 1 s^{-1} , many parallel dislocations gather near the subgrain boundary, as shown in Fig. 3(d). FAN et al [29] reported that the activation of a slip system caused these dislocations, and the higher dislocation density indicated the formation of deformation bands, which would prevent the DRX process. In the Ti-55531 alloys, the reconstructed β phase figure shows the shape of the β grains, the formation of the substructures and the local misorientation. Combined with the grain orientation dispersion (GROD), they believed that DRV followed by CDRX mainly occurred at high strain rates, and only DRV occurred at low and moderate strain rates. The DRV process was described with a Derby relationship of the subgrain size and the β phase stress [54]. In the Ti-1023 alloy, when the strain rate was $4.2 \times 10^{-4} \text{ s}^{-1}$, many LAGBs were produced by DRV at the initial deformation stage ($\varepsilon=0.1$). With the increase in the degree of deformation, the proportion of LAGBs decreased gradually with the occurrence of CDRX [36].

2.2 Dynamic recrystallization behavior and mechanism

Dynamic recrystallization usually interacts with DRV to promote flow softening. The occurrence of DRX needs to meet two necessary conditions of accumulating sufficient dislocation density and reaching the recrystallization temperature. It is mentioned in the introduction that DRX mechanisms can be classified into three categories. It should be noted that one or more DRX mechanisms exist in most β titanium alloys during hot compression [55].

2.2.1 Discontinuous dynamic recrystallization

Generally, if the author does not clearly point out the type of DRX, it generally refers to the traditional DDRX process. DDRX is a typical

strain-induced grain boundary migration (SIGBM) mechanism. DDRX grains always initiate at the grain boundaries because of the high strain gradient. Its high angle grain boundaries (HAGBs) expand and migrate along the dislocation density gradient by absorbing the surrounding substructures and dislocations [56–58]. DDRX usually occurs in metals with low stacking fault energy. This is mainly because the stacking fault energy is low, the extended dislocation width is large, and dislocation cross slipping and climbing are difficult to carry out. As a result, the DRV rate is very low, and the subgrain size is small, which is beneficial to recrystallization nucleation. The critical dislocation density (ρ_c) required to achieve DRX can be expressed by the following equation [59]:

$$\begin{cases} \rho_c = \left(\frac{20\gamma_i \dot{\varepsilon}}{3b^5 l G^2 M_{bm}} \right)^{1/3} \\ M_{bm} = \frac{\beta \delta D_{GB} V_m}{b^2 R} \\ D_{GB} = D_0 \exp \left(-\frac{Q_{diffu}}{RT} \right) \end{cases} \quad (3)$$

where γ_i is the grain boundary energy (J/m^2), l is the subgrain size (nm), M_{bm} is the boundary mobility, $\dot{\varepsilon}$ is the strain rate (s^{-1}), δ is the thickness of the grain boundary (1 nm), D_{GB} is the self-diffusivity along a grain boundary, V_m is the molar volume, D_0 and Q_{diffu} are the self-diffusion coefficient and activation energy, respectively, β is the fraction parameter (estimated to be between 0.1 and 0.5), R is the molar gas constant, and T is the thermodynamic temperature. Equation (3) shows that the deformation conditions significantly affect the critical dislocation density of DRX. The decrease in strain rate leads to a decrease in the critical dislocation density for DRX, which promotes the occurrence of DRX.

Because the stacking fault energy of the BCC structure is high, the width of extended dislocations is narrow, and the climbing and cross slipping of dislocations are easy to carry out, it is not easy to reach the energy needed for the occurrence of DDRX. However, in some studies, DDRX in the β phase region can be observed [29,35,40,60–62]. Its most typical features are a necklace-like grain structure and grain boundary bulge, as shown in Fig. 4(a). In addition, DDRX grains are formed at triple junctions of the original grains [63]. At the

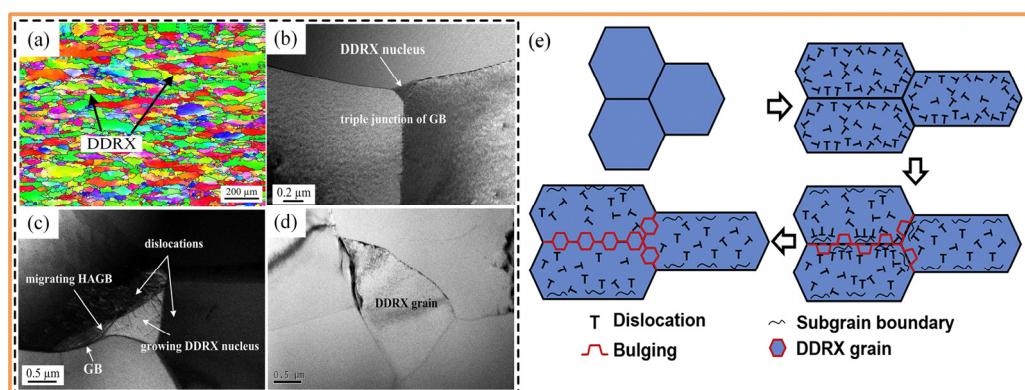


Fig. 4 Typical phenomena and nucleation diagrams of DDRX: (a) Necklace-like grain structure and bulge of grain boundaries [60]; (b) Nucleation at triangular grain boundaries; (c) High dislocation density around DDRX nucleation [35]; (d) Formation of DDRX grains [60]; (e) Schematic diagram of DDRX nucleation process [32]

same time, the dislocation pile-up and dislocation wall are surrounded by the DRX nucleus [60]. When the DDRX grains are formed, no obvious dislocations occur in the DRX grain inside the grains, indicating that DDRX can effectively reduce the dislocation density and eliminate lattice distortion, as shown in Figs. 4(b–d).

Figure 4(e) shows the nucleation mechanism of DDRX, and the typical characteristics are as follows.

(1) With increasing strain, dislocations accumulate at the edge of the original grain boundary. This provides enough substructures and distortion energy to promote the nucleation of DRX grains.

(2) Then, the HAGBs of DRX nuclei start to migrate and extend with the change in dislocation density.

(3) The new DDRX grains engulf the original grains.

2.2.2 Continuous dynamic recrystallization

Many researchers believe that DRV is the main softening mechanism of high stacking fault energy metals. The wide application of EBSD technology in recent years has provided the basis for the discovery of CDRX. For β titanium alloys, the CDRX phenomenon is more common than DDRX. The differences between DDRX and CDRX are mainly in the following aspects [61,63–74].

(1) CDRX is accompanied by the generation of dislocations, DRV and the formation of LAGBs with increasing strain and the formation of many subgrains. Subsequently, CDRX grains occur by progressive rotation of subgrains near the original grain boundary, accompanied by the transformation

of LAGBs into HAGBs (see Figs. 5(a–c)). A gradient misorientation occurs from the center to the boundary in the original grains so that the change in dislocation density inside and outside CDRX grains is smaller than that of DDRX grains (see Figs. 5(d–f)). In addition, the orientation of CDRX grains differs from that of adjoining grains, and subgrains have similar crystallographic orientations.

(2) The kinetics of CDRX is much slower than that of DDRX. It requires a large strain to fully achieve CDRX, but only a small or medium strain is required to activate CDRX.

(3) DDRX is affected by the initial grain size, while CDRX is independent of the initial grain size.

(4) The DDRX has mainly concentrated in β phase region. However, CDRX exists in both the $\alpha+\beta$ phase region and β phase region.

Furthermore, some studies have pointed out that the α phase plays a significant role in CDRX. During the low strain rate deformation of Ti-5553 alloy in the two-phase region, the α/β phase boundaries increase with the increase of α phase precipitation, and the α/β phase boundaries are considered to be the emission sites of movable dislocations, which enhance the CDRX process [43]. However, at deformation temperatures above the phase transition point, DRV becomes the primary softening mechanism due to coarse-grained β . Also, the α phase can promote the transformation from LAGBs to HAGBs [75]. LI et al [34] reported that the CDRX process of the α phase led to the crystal orientation spreading from hard orientation to soft orientation and accelerated the transformation from DRV to CDRX. Meanwhile, the harder α phases

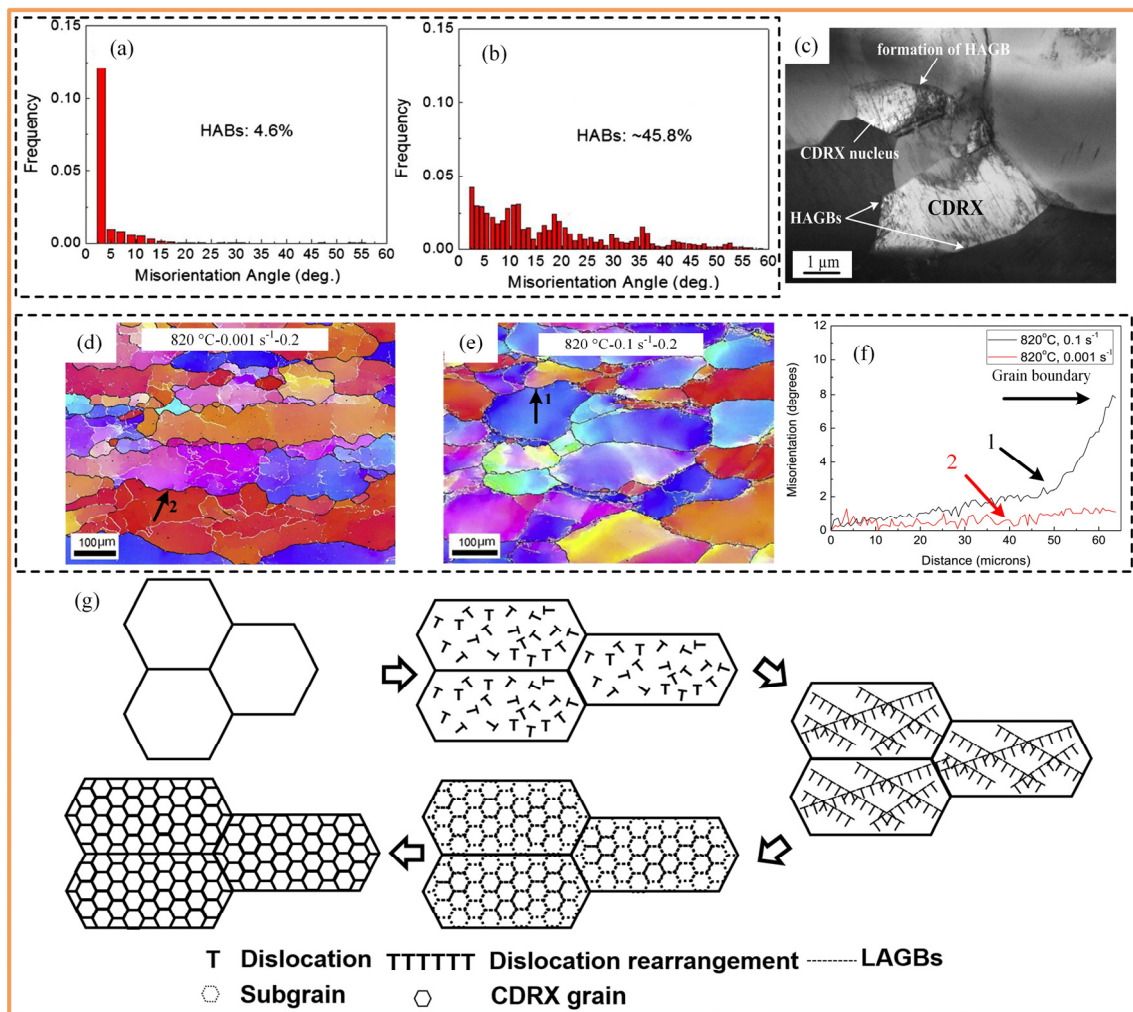


Fig. 5 Typical phenomena and schematic diagrams of CDRX process: (a, b) Distribution of grain boundary misorientation for Ti-52244 alloy at 810 °C and strain rate of 0.01 s⁻¹ to height reductions of 40% (a) and 60% (b) [34]; (c) DRX grain surrounded by distinct HAGBs [35]; (d–f) Misorientation distribution from grain interior to grain boundaries after deformation to strain of 0.2 in Ti-1023 alloy [67]; (g) Schematic diagram of CDRX process

were distributed around the softer β phase, which could inhibit the growth of the β phase and promote the DRX of the α phase and β phase [33].

Figure 5(g) shows a schematic illustration of the CDRX process. In the initial period, a large number of dislocations produced will be rearranged. Then, LAGBs are generated through dislocation rearrangement. Furthermore, LAGBs further absorb dislocations and produce subgrains. As the deformation increases, the accumulation of dislocations leads to the transformation of LAGBs into HAGBs. It is worth noting that the formation of subgrains in the initial deformation stage is considered to be completed by DRV. On the other hand, FAN et al [29] still found an obvious CDRX process in the β phase region at a high strain rate,

and they also proposed a two-step CDRX behavior, which is different from the traditional CDRX behavior. They believe that many substructures are mainly distributed on one side of the β grain boundary due to heterogeneous plastic deformation. With the increase of strain, the substructures merge, and new HAGBs are produced. Simultaneously, ribbon grain structures are formed at the original boundaries. Subsequently, the substructure in the ribbon grain structure migrates to form HAGBs driven by the locally high stored energy. Finally, HAGBs gradually separate the ribbon grain structure into smaller grains [29]. However, the orientation of grain boundary accumulation in these two steps of CDRX is similar to that of traditional CDRX.

2.3 Dynamic behavior of α phase during deformation in two-phase region

2.3.1 Dynamic phase transformation

The interaction between deformation and phase transformation easily occurs when titanium alloy is hot deformed in the two-phase region. Dynamic phase transformation is used to study the influence of deformation parameters on transformation characteristics, including strain-induced martensite transformation, $\alpha \rightarrow \beta$ transformation and α phase precipitation ($\beta \rightarrow \alpha$). This article mainly reviews the latter two transformation behaviors.

The deformation stress can provide the driving force for the phase transformation. In addition, the diffusion conditions and distribution characteristics of solute elements during deformation are different from those without deformation, which leads to a change in the two-phase equilibrium state. The dynamic phase transformation was firstly found in Fe–Ni–C alloys. YADA et al [76] found the transformation of $\gamma \rightarrow \alpha$ in the hot torsion process. At the same time, dynamic phase transformation was found in pure titanium [77], Ti–6Al–4V [78,79], and Ti–Al–Fe [80], and the critical conditions, driving forces and effects of dynamic phase transformation on flow softening were analyzed. Furthermore, phase transformation serves as another kind of deformation mechanism for crystal plasticity [81], including α phase precipitation behavior in the following article. It should be pointed out that the initial microstructure of the dynamic phase transformation is a mixed microstructure of the α phase and β matrix, which needs to be deformed in the two-phase region.

Many studies have considered that the deformation conditions significantly affect the dynamic phase transformation. LI and YANG [82] studied the strain-induced α -to- β phase transformation of Ti-55511 alloy. Compared with undeformed samples, the volume fraction of the α phase at different deformation temperatures was calculated by Thermal-Cale, as shown in Fig. 6. Under the same conditions, the α phase in the deformed sample decreased with increasing temperature compared with that in the undeformed samples. They thought that the stress was concentrated near the phase boundary when the sample was deformed from the thermodynamic point. The stress

concentration would change the total free energy of the sample. Therefore, the phase system equilibrium near the boundaries was broken, and the transformation from α into β could be promoted under compressive stress. The phase transformation process was mainly controlled by the diffusion of solute elements in the deformed state. From the perspective of dynamics, the diffusion of solute elements was the condition for strain-induced phase transformation. The defects induced by deformation provided diffusion channels for solute elements, and the deformation storage energy provided power for diffusion near the interface. This led to a redistribution of solute elements and provided an additional driving force for the transformation of α into β [82]. LIN et al [33] also reported a similar phenomenon. They found that the effects of strain and strain rate on the dynamic phase transformation are caused by an increase in dislocation density. Furthermore, AN et al [83] found that deformation heating at a high strain rate resulted in the transformation of the fine second phase α into the β phase in Ti-555211 alloy.

WU et al [84] analyzed the dynamic phase transformation behavior of Ti-55531 alloy in the $\alpha + \beta$ phase region. When the strain rate and

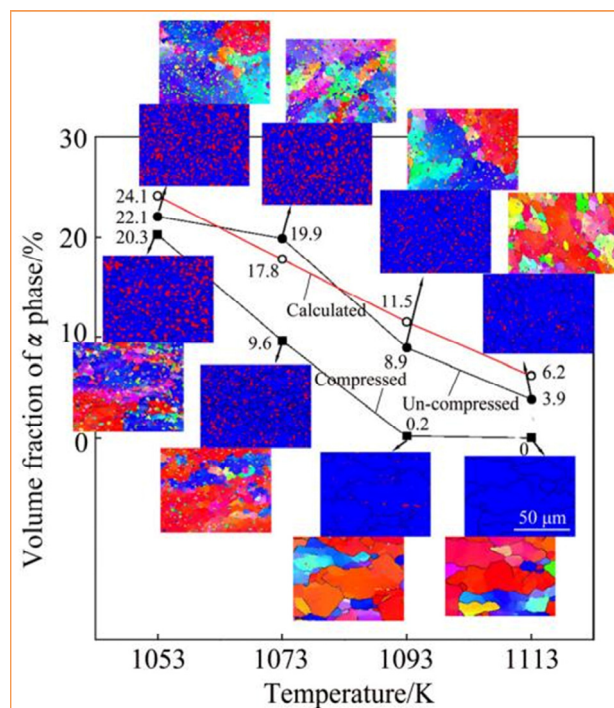


Fig. 6 Phase distribution and orientation maps of both compressed and uncompressed samples at 1053, 1073, 1093 and 1113 K (Deformed samples are compressed with 0.01 s^{-1} by reduction of 60%) [82]

temperature were constant, with the continuous increase in deformation, an obvious α -to- β phase transformation occurred in the sample, and the α phase proportion decreased from 35% to 23%. Furthermore, according to the difference between peak stress and yield stress to calculate the driving force of phase transformation [79], they found that the driving force ranges from 5 to 170 MPa. Therefore, thermo-mechanical processing can effectively improve the phase transformation kinetics [85]. The change in stress caused by the dynamic phase transformation can be calculated by the following formula [79]:

$$\Delta\sigma_{\text{DPT}} = \Delta V_{\text{DPT}} (M_{\alpha}\tau_{\alpha} - M_{\beta}\tau_{\beta}) \quad (4)$$

where $\Delta\sigma_{\text{DPT}}$ denotes the stress change due to the dynamic phase transformation and ΔV_{DPT} denotes the change of the volume fraction of the α phase due to the dynamic phase transformation. τ_{α} and τ_{β} are the critical resolved shear stresses of the α phase and β phase, respectively, and M_{α} and M_{β} are the corresponding Taylor factors, respectively. The softening fraction by dynamic phase transformation is defined as $f_{\text{DPT}} = \Delta\sigma_{\text{DPT}}/\sigma_p$. By calculating the softening fraction of Ti-55531 alloy, the dynamic phase transformation played a main role at a strain rate lower than 0.1 s^{-1} , and the softening fraction was in a range from 14.7% to 20% [84].

In β titanium alloy, the influence of $[\text{Mo}]_{\text{eq}}$ and element diffusion result in the fact that the precipitation of the α phase is more sensitive than that of the $\alpha+\beta$ titanium alloy. The additional alloy elements (such as Al, Cr, Nb and V) affect α phase precipitation [86]. The existing research mainly focuses on the orientation change, size, morphology and volume fraction of the precipitation phase. According to the Burgers orientation relationship (BOR), the orientation variants result from β -to- α phase transformation, as shown in Table 1 [87]. Unlike the dynamic phase transformation, α phase precipitation occurs when the initial misrostructure is the β phase and deforms in the two-phase region. Four types of orientation occur during the hot compression of Ti-55511 alloy, as shown in Fig. 7. The first type is a typical BOR (Fig. 7(c)). The second type shares the $\langle 11\bar{2}0 \rangle$ axis with that of the BOR type (Figs. 7(b, d)). Figure 7(e) shows the relationship between the second type and BOR (Fig. 7(f)). The third type is the derivative one from

BOR. The orientation relationship between the third type and β matrix is $\{0001\}_{\alpha} // \{111\}_{\beta}$ and $\langle 10\bar{1}0 \rangle_{\alpha} // \langle 110 \rangle_{\beta}$. The fourth type has the same c -axis and 30° rotation angle as the third type (Fig. 7(g)). Figure 7(h) shows the orientation relationship between the third type and the fourth type. Simultaneously, Fig. 7(i) shows that α precipitates without an orientation relationship with the β phase can be observed. They can be regarded as the random orientation [39]. Simultaneously, in Fig. 7(a), α precipitates are densely distributed with few orientations near the β grain boundary compared to that in the β grain interior.

However, no obvious BOR was found in the α phase precipitation during hot deformation of the Ti-7333 alloy, but the pole figure of α phase at the grain boundaries (GBs) shows that there was a near BOR between the α and β phases. The main reason was that the rotation and globularization process of the α phase broke the BOR. The precipitation phase was also concentrated near the grain boundary, and a small part was concentrated in the grain interior [86,88]. These studies further illustrate the random orientation of α precipitates.

In addition, the size of the precipitation phase is significantly affected by the deformation parameters. HUA et al [86] found that the size of the grain boundaries and intragranular α precipitates during hot compression was much smaller than that of α precipitates during heat treatment. They

Table 1 Orientation variants resulting from β -to- α phase transformation according to Burgers relationship [87]

Variant No.	Corresponding plane, $(hkl)_{\beta} // (hkl)_{\alpha}$	Corresponding direction, $[uvw]_{\beta} // [uvw]_{\alpha}$
1	$(001)_{\beta} // (0001)_{\alpha}$	$[\bar{1}\bar{1}1]_{\beta} // [2\bar{1}\bar{1}0]_{\alpha}$
2		$[1\bar{1}1]_{\beta} // [2\bar{1}\bar{1}0]_{\alpha}$
3	$(001)_{\beta} // (0001)_{\alpha}$	$[1\bar{1}1]_{\beta} // [2\bar{1}\bar{1}0]_{\alpha}$
4		$[111]_{\beta} // [2\bar{1}\bar{1}0]_{\alpha}$
5	$(0\bar{1}1)_{\beta} // (0001)_{\alpha}$	$[111]_{\beta} // [2\bar{1}\bar{1}0]_{\alpha}$
6		$[\bar{1}11]_{\beta} // [2\bar{1}\bar{1}0]_{\alpha}$
7	$(101)_{\beta} // (0001)_{\alpha}$	$[\bar{1}11]_{\beta} // [2\bar{1}\bar{1}0]_{\alpha}$
8		$[\bar{1}\bar{1}1]_{\beta} // [2\bar{1}\bar{1}0]_{\alpha}$
9	$(\bar{1}10)_{\beta} // (0001)_{\alpha}$	$[\bar{1}\bar{1}1]_{\beta} // [2\bar{1}\bar{1}0]_{\alpha}$
10		$[111]_{\beta} // [2\bar{1}\bar{1}0]_{\alpha}$
11	$(110)_{\beta} // (0001)_{\alpha}$	$[1\bar{1}1]_{\beta} // [2\bar{1}\bar{1}0]_{\alpha}$
12		$[\bar{1}11]_{\beta} // [2\bar{1}\bar{1}0]_{\alpha}$

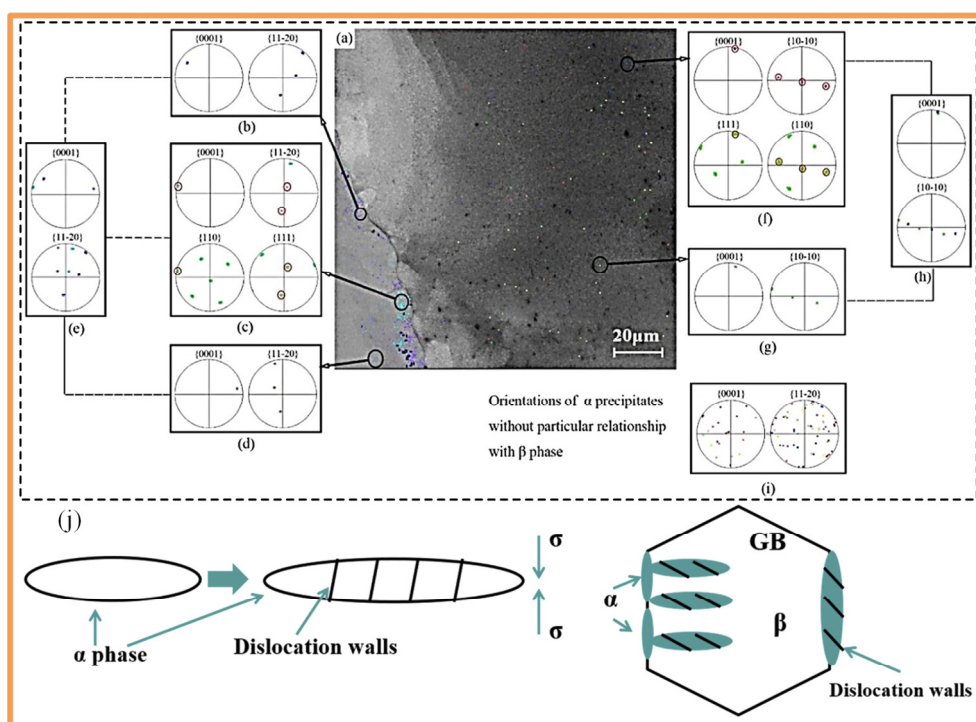


Fig. 7 Orientation types of α precipitates during hot compression: (a) Geometrical distribution of orientations; (b) One orientation of the second type; (c) BOR type; (d) Other orientation of the second type; (e) Orientation relationship of BOR type and the second type; (f) The third type; (g) The fourth type; (h) Orientation relationship of the third type and the fourth type; (i) Random type (Dotted lines indicate the orientation relationships of linked orientations) [39]; (j) Schematic illustration of α phase precipitation behavior [86]

believed that the heat treatment process was an element diffusion process, while the precipitation in the compression deformation was controlled by the diffusion and the force simultaneously. Moreover, spherical α phase precipitation was beneficial to the plasticity of the alloy. The above research only studied the precipitation behavior of the α phase under single conditions without considering the influence of deformation parameters. Therefore, WANG et al [39] found that the content, aspect ratio and size of the precipitated α phase decreased with increasing temperature, but increased firstly and then decreased with increasing strain rate during hot deformation of Ti-55511 alloy. This provided important guidance for controlling the morphology of the α phase in industrial production [39]. The α phase precipitation mechanism is shown in Fig. 7(j). The defect structure and stress state control the α phase precipitation during the deformation process. The deformation process significantly increases the proportion of LAGBs, skewing the misorientation-angle distribution towards low-angle rotations.

LAGBs are used as nucleation sites, and similar LAGBs angles lead to the approximate crystal orientation of the precipitates [89]. Meanwhile, defect coupling with stress leads to a fine equiaxed morphology [90]. During deformation, many dislocations around the α phase lead to stress concentrations. Therefore, shearing strain in the α phase is caused by large deformation, the generation of dislocations and the nucleation of the dislocation interface in the α phase. Finally, dislocations lead to precipitates of different morphologies and sizes [86].

2.3.2 Kinking and globularization of α phase

The dynamic behavior of the α phase is a typical phenomenon in the hot deformation process of titanium alloy in the two-phase region. The dynamic behavior includes kinking (Fig. 8(a)), buckling, fragmentation (Fig. 8(b)), and globularization (Fig. 8(c)), mainly to relax the stress concentrations during deformation and promote grain refinement. They mainly appear in the hot deformation process of the lamellar α phase. Lamellar α is a typical microstructure of titanium

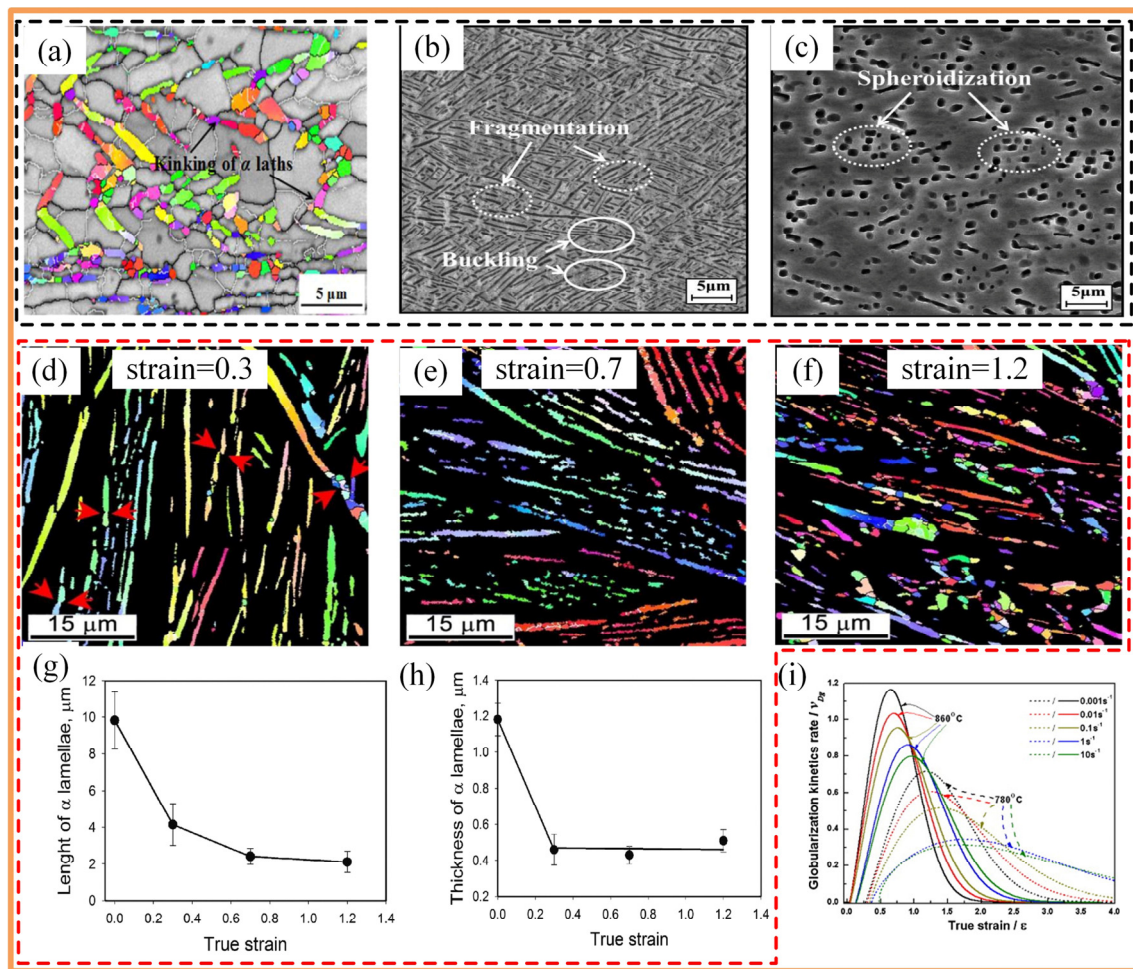


Fig. 8 Dynamic behavior of lamellar α : (a) Kinking [91]; (b) Buckling and fragmentation; (c) Globularization [64]; (d–f) α phase at different strains of 0.3 (d), 0.7 (e) and 1.2 (f); (g) Variation of length of lamellar α phase with strain in (d–f); (h) Variation of thickness of lamellar α phase with strain in (d–f) [92]; (i) Variation of dynamic globularization kinetic rate under different deformation conditions [93]

alloys, in which lamellar α is always formed as many colonies within the former β grains [28]. The dynamic behavior is closely affected by the deformation temperature, strain rate, and strain. The extent of kinking of the lamellar α phase is related to the deformation parameters and its distribution direction and size. This distribution direction will affect the degree of difficulty of dynamic globularization during the deformation process. Generally, buckling and kinking are more likely to occur when the distribution direction is close to the compression direction. When the lamellar α phase is perpendicular to the compression direction, it is more likely to buckle or even fragment due to the shear force. The increase in deformation temperature can reduce the deformation resistance, thus decreasing the proportion of the α phase. The

lamellar α phase is more likely to kink at a high strain rate [94]. A higher temperature and lower strain rate will promote the transformation of buckling into fragmentation [95]. In addition, kinking of the lamellar α phase can promote the CDRX process of the β phase, and the formation of globular α assists in DRX of the β phase through grain boundary sliding [34]. The new study shows that the accumulation of dislocations at the α/β interface can induce the formation of twins. The twin boundary migration leads to the formation of a substructure in the acicular α , followed by meeting another adjacent parallel twin and then causing twin merging. The accumulation of twin displacement leads to the bending of acicular α [96].

Among all the factors that affect dynamic globularization, the deformation amount is the

most influential factor. Many studies have pointed out that lamellar α kinking usually occurs at low strain [52,75,93,94]. Moreover, the kinking process of the lamellar α phase is accompanied by obvious crystallographic rotation [97], resulting in a certain number of LAGBs and HAGBs, and then these HAGBs can further promote the globularization of lamellar α [75]. Based on the microstructures of the core, middle, and edge, JONES et al [98] and JACKSON et al [99] further illustrate the effect of strain on the dynamic globularization of the α phase. Figures 8(g, h) show the change in the length and thickness of the lamellar α phase with strain in Figs. 8(d–f). The length and thickness of the lamellar α decrease significantly with increasing strain, which is directly related to the kinking and fragmentation of the α phase. In addition, WANG et al [93] also established the functional relationship between the dynamic globularization kinetic rate and strain, as shown in Fig. 8(i). The main reason for the increase of dynamic globularization kinetic rate at first is the breakup of a large amount of abundant prior lamellar α , while the decrease in the dynamic rate at large strain may be due to the absence of prior lamellae and difficulty of unceasing fragmentation of the broken lamellar α phase. Simultaneously, the existence of globularized α phase under large strain reduces the globularization kinetic rate [100]. The globularization kinetic rate at a low strain rate is much higher than that at a high strain rate. This is due to the short deformation time at higher strain rates, and the α phase does not have enough time for dynamic recovery and recrystallization [101]. The increase of temperature and the decrease of strain rate will increase the maximum kinetic rate, indicating that the diffusion process plays an essential role in the globularization process.

It is worth noting that the dynamic globularization mechanism is one of the main hotspots in the research of β titanium alloys. In these studies, the dynamic globularization mechanism in β titanium alloy is basically divided into two categories: one is grain boundary separation and wedging of the β phase [33,44,64,71,95,102–106]. During the deformation process, subgrain boundaries are formed in the lamellar α phase, and the subgrain boundaries will slip under the action of shear force and produce unstable interfaces. An unstable

internal interface is formed within the lamellar structure, and its dihedral angle is 90° . To reduce the surface tension, the dihedral angle will gradually decrease. Furthermore, some studies have pointed out that this process is accompanied by the formation of a thermal erosion ditch and element diffusion [35,85,95,106,107]. Saturated elements of V, Mo and Cr will diffuse from the α phase to the adjacent β phase through the thermal erosion ditch, leading to the β phase wedging into α and separating the α precipitates into two parts. In the meantime, the Al and O elements in the nonequilibrium β matrix will migrate towards the α precipitates, resulting in the dynamic coarsening of the α phase and the change in grain orientation. These two processes are jointly driven and promoted by the element concentration gradient and the external driving force introduced by hot deformation. Finally, the α/β interface migrates gradually to form a globularized structure, as shown in Figs. 9(a, b). The occurrence of a phase boundary is the first step in the breakup of the lamellar α phase [85,108]. When the width of the lamellar α is less than two times the penetration distance of the β phase cusps, the separation of primary α grains occurs readily [108]. The other is the shearing of the lamellar α phase [38,86,109,110], which mainly includes four steps: lath shearing, dislocation generation, nucleation of the dislocation interface and interface migration [111], as shown in Figs. 9(c, d). Firstly, under the action of shear strain, the lamellar α phase undergoes shear deformation, and the deformation amount controls the shear deformation process. In hot deformation, opposite sign dislocations are formed in the α phase. Simultaneously, the high dislocation density also leads to the obvious DRV in the deformation process, which promotes the annihilation of the opposite sign dislocations on the slip plane. Subsequently, the opposite sign dislocations will form a new α/β interface along the shear plane. Compared with the β phase wedge model, α phase shear is formed under a strong local stress concentration without the redistribution of alloying elements. ZHAO et al [60] found that the lamellar α in PM Ti-5553 alloy is directly sheared into two parts. The globularization mechanism of β titanium alloy also includes the termination migration mechanism [92].

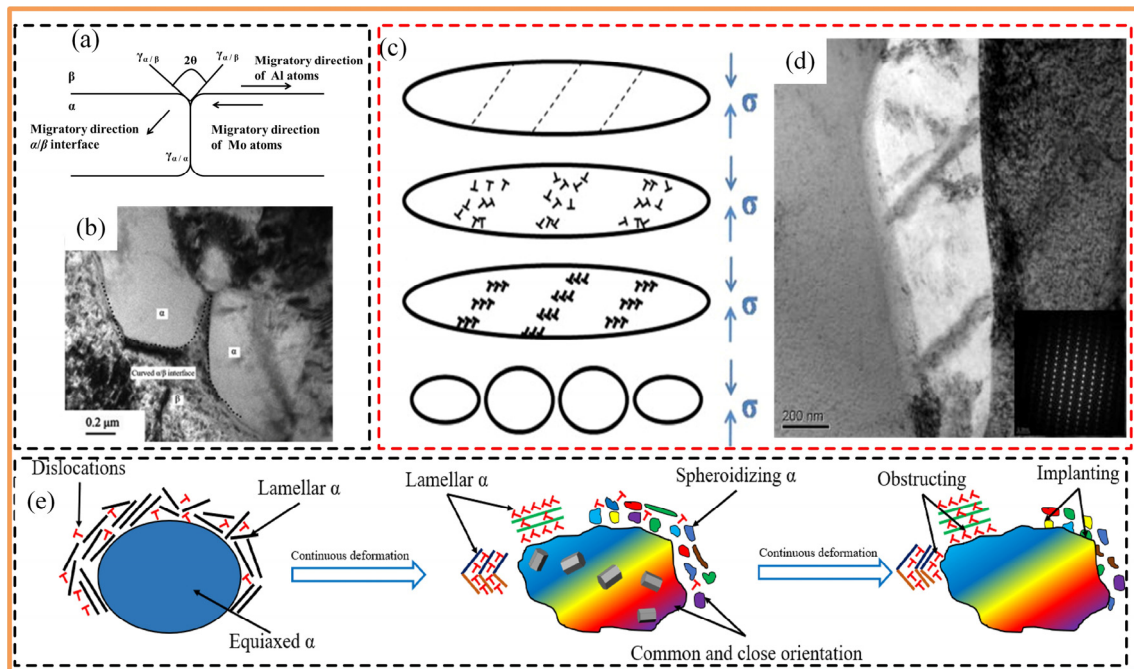


Fig. 9 Two main globularization mechanisms and TEM characterization of lamellar α phase in β titanium alloy: (a, b) β phase wedge model [85] and TEM characterization [95]; (c, d) α phase shear mechanism and TEM characterization [38,109]; (e) Implantation mechanism of lamellar α phase [112]

It is worth noting that the globularization of the α phase includes the progressive formation of LAGBs via DRV, and with further deformation, LAGBs are transformed into HAGBs [75,97]. For different materials, the globularization mechanism is not single. Two globularization mechanisms occur in Ti-52244 alloys [65], PM Ti-5553 alloys [60] and Ti-6554 alloys [113].

In the latest study, LIN et al [112] proposed a new globularization mechanism in Ti-55511 alloy with initial equiaxed α and lamellar α , as shown in Fig. 9(e). Firstly, dislocations begin to accumulate in the early stages of deformation. The rotation, bending and rearrangement of the lamellar α phase lead to different orientations. Meanwhile, the lamellar α phase is difficult to implant into the equiaxed α phase because of dislocations moving between the lamellar α and equiaxed α . Finally, the lamellar α phases are completely globularized, and the globularized α phases near the equiaxed α phases gradually form common orientation relationships. Globularized α phase was implanted into the equiaxed α phase.

2.4 Orientation transformation and texture evolution

The formation of texture in titanium alloy is

affected by DRX and dynamic phase transformation. The current research on the crystal orientation and texture of titanium alloy mainly focuses on the orientation relationship between the α and β phases and deformation parameters on the texture strength and composition. Many studies have pointed out that hot deformation changes the strict BOR between the α phase and β phase [34,38,45,71,88,92,114,115] (Fig. 10(c)), and a high strain rate accelerates the loss of coherency of BOR at the α/β interface, as shown in Figs. 10(f, i). In addition, SUN et al [71] found that the strain value deviating from BOR was 0.36, and the loss of coherence between α and β interfaces was always accompanied by the development of LAGBs and HAGBs. Moreover, KLIMOVA et al [92] also found that when the true strain was 1.2, the average deviation from BOR reached 34.5° , and the interphase boundaries changed from semi-coherent to non-coherent semicoherent and then to noncoherent. HUA et al [81], based on the dislocation movement and β to α phase transformation, found that in the steady-state stage, the deformation behavior of β grains could be further divided into two types, i.e., Type I with its $\{110\}_\beta \langle 1\bar{1}1 \rangle_\beta$ slip system in a favourable orientation relative to the external compressive load

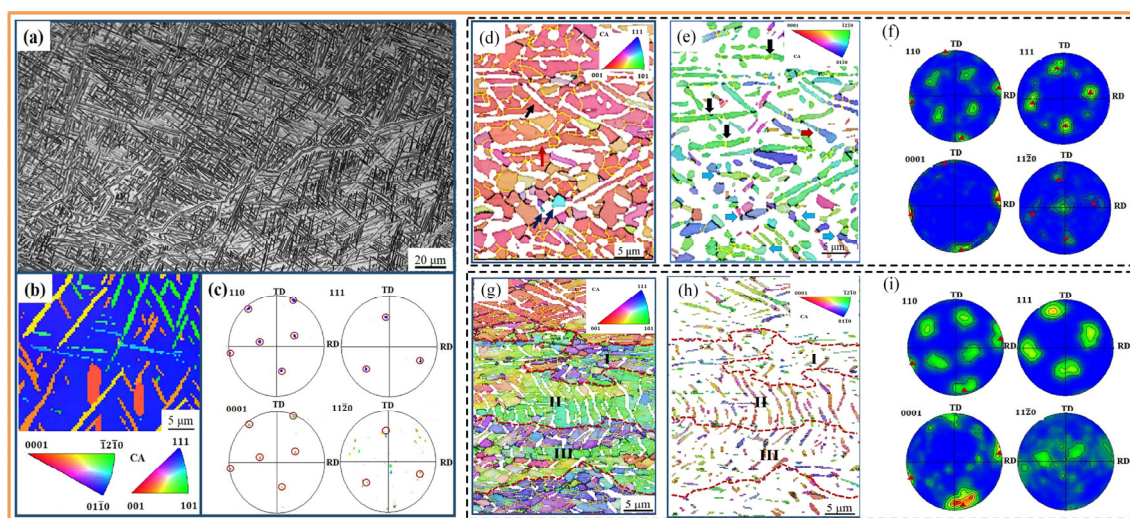


Fig. 10 Initial microstructures of Ti-52244 alloy: (a) Optical micrograph, (b) inverse pole figure ($//CA$) and (c) pole figures; Microstructures at 1073 K and strain rates of 0.0002 s^{-1} to strain of 0.36: (d) β phase, (e) α phase, and (f) pole figures; Microstructure at 1073 K and strain rates of 0.1 s^{-1} to strain of 0.36: (g) β phase, (h) α phase, and (i) pole figures [71]

and Type II with its $\{1\bar{1}2\}_{\beta}\langle\bar{1}1\bar{1}\rangle_{\beta}$ slip system in a favourable orientation. Their deformation processes are realized by the formation of dislocations and intragranular α precipitates, respectively.

Plastic deformation leads to an increase in the dislocation density, while recovery and recrystallization lead to a decrease in the dislocation density, and their interaction will affect the types and strength of the deformation texture and recrystallization texture. LI et al [75] found that the $\langle 001 \rangle$ fiber texture was formed due to DRV at a low strain rate, which led to a decrease in the deformation texture strength. With increasing strain rate, CDRX weakened the deformation texture. CDRX could cause a large misorientation between the newly formed recrystallized grains, which led to a decrease in the overall texture strength. LI et al [75] also pointed out that the degree of contribution of DRV and DRX to softening also led to differences in texture types and strengths. Some studies have pointed out that DRX can enhance the $\{100\}$ texture [38,45,51]. In a follow-up study, LI and YANG [116] found that the dynamically induced grain boundary migration under compressive stress was the main reason for $\langle 100 \rangle$ oriented grains to engulf $\langle 111 \rangle$ oriented grains, resulting in the strengthening of the $\langle 100 \rangle$ texture and the weakening of the $\langle 111 \rangle$ texture. In the two-phase region, the α phase precipitation showed

a strong $\{11\bar{2}0\}$ phase transformation texture. The reason was that the $\{10\bar{1}1\}\langle 11\bar{2}0 \rangle$ slip system enhances the slip transmission between the α and β phases and the $\{110\}\langle 111 \rangle_{\beta}$ slip activity [51].

In addition, the deformation amount also has a significant effect on texture evolution. WU et al [84] found that a Ti-55531 alloy with a strong initial texture firstly decreased and then increased during deformation in the two-phase region. They thought that the initial orientation was broken due to the rotation and deformation of all grains. With increasing deformation, most grains rotate towards a preferential direction, enhancing the texture strength [84]. A similar trend of texture strength change also appears in the Ti-52244 alloy, but the author thinks that the decrease in texture strength is caused by the kinking of the lamellar α phase [34]. Some studies have also pointed out that the strength of fiber texture $\langle 100 \rangle$ decreases with increasing strain rate [38,45], which further shows that the deformation parameters will also affect the texture evolution. Furthermore, the cooling rate also has a significant effect on the texture. CHEN et al [114] found that $\{111\}\langle 110 \rangle$ and $\{111\}\langle 112 \rangle$ weak γ fiber textures are formed at a fast cooling rate. According to the above research, the texture change is significantly affected by the deformation conditions and the initial microstructure of the sample, and the softening mechanism also plays an important role.

3 Plastic deformation processing map

3.1 Typical β titanium alloy processing map

The hot processing map can optimize the process parameters, reveal the plastic deformation mechanism, and control the microstructure and mechanical properties. At present, the most common method to establish hot processing map is the dynamic material model (DMM). According to this model, the energy input by external forces to the material is mainly used for plastic deformation and microstructure evolution, in which most of the plastic deformation is converted into heat and a small part is stored in the sample. The relevant parameters in the hot processing map are calculated by the following formula [117,118]:

$$m = \frac{d(\lg \sigma)}{d(\lg \dot{\epsilon})} = b_1 + 2c \lg \dot{\epsilon} + 3d(\lg \dot{\epsilon})^2 \quad (5)$$

$$\eta = \frac{2m}{1+m} \quad (6)$$

$$\xi(\dot{\epsilon}) = \frac{\partial \lg[m/(m+1)]}{\partial \lg \dot{\epsilon}} + m < 0 \quad (7)$$

where b_1 , c , and d are material constants, η is the efficiency of power dissipation, m is the strain rate sensitivity coefficient and $\xi(\dot{\epsilon})$ is the rheological instability parameter. Using polynomial fitting can obtain the value of m and then obtain the value of η and $\xi(\dot{\epsilon})$. Many studies have pointed out that strain significantly influences the power dissipation map and the instability map [22,37,53,119–121]. This is mainly because the m value is significantly affected by the deformation parameters. This article summarizes the optimal hot processing region and instability region under different deformation parameters of β titanium alloy in recent years, as shown in Table 2. Figure 11 establishes the optimal processing regions and instability regions of different alloys according to Table 2. The optimal hot processing regions mainly appear in the low strain rate region, and instability regions mainly appear in the high strain rate region. This shows that the high strength β titanium alloy is suitable for hot working at a low strain rate. The flow stability mainly includes dynamic globularization, dynamic recrystallization and superplastic deformation. The optimal working area has a higher energy dissipation efficiency. The peak efficiency of power

dissipation occurs not only in the area where the strain rate is low but also appears near the phase transition point. This phenomenon was found in Ti-55511 alloys [53] and Ti-5553 alloys [60]. The reason is the $\alpha \rightarrow \beta$ phase transformation. Among them, the hot processing map of the Ti-5553 alloy produced instability in the middle strain rate region, and the author thinks that this phenomenon was caused by the overlapping of stress–strain curves of strain rates of 1 and 0.1 s^{-1} due to dynamic α precipitation [43]. Considering that strain has a more significant influence on the hot processing map, the three-dimensional hot processing map established by superimposing the hot processing under different strains can more clearly reflect the effect of strain on the hot workability. This method was used in TC17 alloy [94].

3.2 Relationship between efficiency of power dissipation and deformation mechanism

The efficiency of power dissipation (η) plays a vital role in the hot processing map. The η value reflects all kinds of microstructure change mechanisms in a specific range of deformation temperatures and strain rates. According to the existing research, it is concluded that the efficiency value and the deformation mechanism are as follows [111,132,133].

(1) Low efficiency from 0.1 to 0.3 corresponds to the DRV process.

(2) High efficiency from 0.35 to 0.55 corresponds to the DRX process.

(3) Higher efficiency up to 0.7 corresponds to superplasticity.

According to the above theory, LIANG et al [134] calculated the efficiency of power dissipation of Ti-55511 alloy at true strains of 0.1 and 0.2. The areas with higher efficiency were mainly concentrated in areas with lower strain rates. Due to the non-distortion nucleation and subgrain migration, DRX was considered an energy-cost process. Simultaneously, they combined with the OM characterization and confirmed that obvious DRX occurred in the higher efficiency region, further indicating that the efficiency and deformation mechanism conformed to the above relationship. Based on this research, NING et al [135] established the DRX softening map of Ti-55511 alloy based on the efficiency of power dissipation, thermal activation energy, and

Table 2 Optimal hot processing region and instability region of β titanium

Typical β titanium alloy	Deformation parameter	Optimal hot processing region	Instability region
Ti-1023 [119]	675–850 °C, 0.001–0.01 s ⁻¹	(1) 675–850 °C, 0.001–0.01s ⁻¹ (2) 750–850 °C, 0.01–0.1s ⁻¹ (3) 763–804 °C, 0.1–1 s ⁻¹	675–727 °C, 1–10 s ⁻¹
Ti-55511 [53]	720–930 °C, 0.01–5 s ⁻¹	(1) 720–756 °C, 0.01–0.046s ⁻¹ (2) 888–930 °C, 0.059–0.21 s ⁻¹	720–930 °C, 1–5 s ⁻¹
Ti-5553(PM) [60]	800–1150 °C, 0.001–10 s ⁻¹	910–1130 °C, 0.001–0.1s ⁻¹	800–1025 °C, 1–10 s ⁻¹
Ti-5553(IM) [43]	600–1100 °C, 0.001–10 s ⁻¹	(1) 700–780 °C, 0.001–0.0056 s ⁻¹ (2) 840–1000 °C, 0.004–0.0056 s ⁻¹ (3) 600–1100 °C, 0.05–1s ⁻¹	600–1100 °C, 0.05s ⁻¹ –1s ⁻¹
Ti-55531 [122]	860–900 °C, 0.001–10 s ⁻¹	860–900 °C, 0.001–0.0025 s ⁻¹	860–900 °C, 1–10 s ⁻¹
Ti-555211 [123]	750–950 °C, 0.001–1 s ⁻¹	(1) 800–850 °C, 0.001–0.01 s ⁻¹ (2) 900–950 °C, 0.001–0.01 s ⁻¹	(1) 750–825 °C, 0.4–1 s ⁻¹ (2) 925–950 °C, 0.25–1 s ⁻¹
Ti-7333 [109]	770–970 °C, 0.001–10 s ⁻¹	840–880 °C, 0.001–0.003 s ⁻¹	(1) 770–970 °C, 1–10 s ⁻¹ (2) 770–790 °C, 0.0025–0.056 s ⁻¹
Ti-6554 [112]	690–840 °C, 0.001–1s ⁻¹	767–840 °C, 0.001–0.003s ⁻¹	690–822 °C, 0.056–1 s ⁻¹
Ti-4Al-1Sn-2Zr- 5Mo-8V-2.5Cr [124]	780–930 °C, 0.001–1 s ⁻¹	780–880 °C, 0.001–0.01s ⁻¹	780–800 °C, 0.316–1s ⁻¹
Ti-15-3 [125]	850–1150 °C, 0.001–10 s ⁻¹	(1) 925–975 °C, 0.001 s ⁻¹ (2) 1050–1100 °C, 0.001 s ⁻¹	850–1140 °C, 0.1–10 s ⁻¹
TB8 [126]	850–1000 °C, 0.001–1 s ⁻¹	880–985 °C, 0.00191–0.001 s ⁻¹	850–1000 °C, 0.032–1 s ⁻¹
TB9 [46]	800–1000 °C, 0.001–10 s ⁻¹	1000 °C, 0.5 s ⁻¹	800–1000 °C, 1.7–10 s ⁻¹
TB17 [22]	775–905 °C, 0.001–10 s ⁻¹	(1) 795–810 °C, 0.001 s ⁻¹ (2) 865–885 °C, 0.001 s ⁻¹	775–900 °C, 0.1–1.7 s ⁻¹
TB19 [127]	700–1000 °C, 0.001–10 s ⁻¹	(1) 700–800 °C, 0.001–0.1 s ⁻¹ (2) 800–1000 °C, 0.001–0.1 s ⁻¹	(1) 700–800 °C, 0.1–10 s ⁻¹ (2) 800–1000 °C, 1–10 s ⁻¹
Ti-1300 [128]	800–890 °C, 0.001–10 s ⁻¹	(1) 800–820 °C, 0.01–0.05 s ⁻¹ (2) 840–870 °C, 0.01–0.1 s ⁻¹	840–890 °C, 1–10 s ⁻¹
Ti-10V-4.5Fe- 1.5Al [129]	680–900 °C, 0.001–100 s ⁻¹	850 °C, 0.01 s ⁻¹	700–900 °C, 10–100 s ⁻¹
Ti-6246 [130]	800–1000 °C, 0.001–100 s ⁻¹	(1) 800–900 °C, 0.01–0.04 s ⁻¹ (2) 1075–1100 °C, 0.01–0.02 s ⁻¹	800–950 °C, 0.04–4 s ⁻¹
Ti-3.0Al-3.5Cr- 2.0Fe-0.1B [131]	800–950 °C, 0.01–10 s ⁻¹	880–940 °C, 0.01–0.03 s ⁻¹	850 °C, 10 s ⁻¹

Poliak–Jonas criterion and compared with the microstructural observation. It was found that there was a certain difference between the deformation mechanism obtained by the above three methods and the microstructural observation. However, the power efficiency map had the highest similarity with the microstructural observation. FAN et al [29] established the relationship between the efficiency of power dissipation and the deformation mechanism of Ti-55531 alloy, as

shown in Fig. 12(a). The deformation mechanism in the high efficiency region is mainly the DDRX process. With the increase of strain rate, CDRX and deformation bands (DBs) occur. When the strain rate reached 10 s⁻¹, the deformation mechanism was mainly DBs and shear bands (SBs), and DBs and SBs could hinder the DRX process. Furthermore, the occurrence of CDRX at low strain rates also had higher efficiency during the hot deformation of the Ti-5553 alloy. This was mainly due to the increase

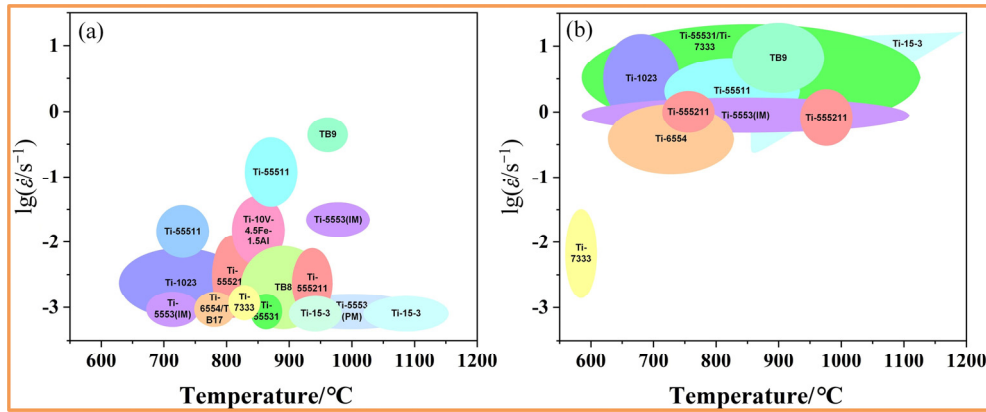


Fig. 11 Optimal hot processing regions (a) and instability regions (b) of typical β titanium alloy

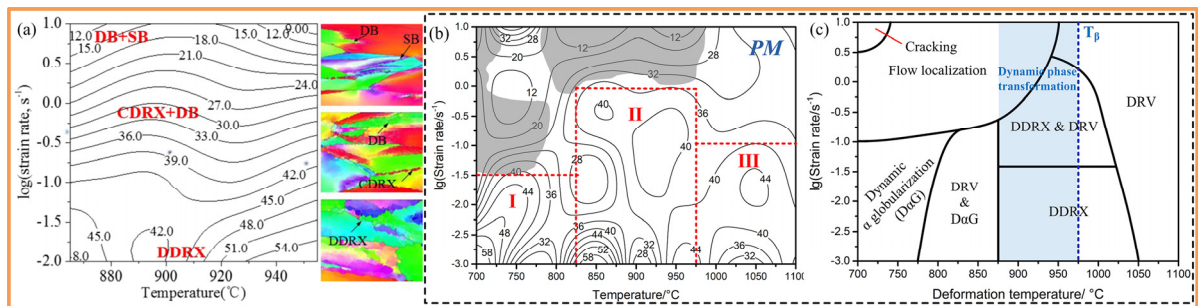


Fig. 12 Relationship between efficiency of power dissipation and deformation mechanism for Ti-55531 alloy (a) [29], and PM Ti-5553 alloy (b, c) [35]

at α/β boundaries accompanied by α precipitation, leading to an increase in emission sites of moving dislocations. Therefore, it promoted the CDRX process [43]. ZHAO et al [60] refined the relationship between the efficiency of power dissipation and the deformation mechanism in PM Ti-5553 alloy, as shown in Figs. 12(b, c). They found that dynamic α globularization and superplastic deformation were dominant in high power efficiency areas. Meanwhile, they carried out a high temperature tensile test in the peak efficiency area and found that the elongation of Ti-5553 alloy could reach 103.5%. With the decrease of power efficiency, the softening mechanism gradually changed to DDRX and DRV. Similar methods are also used in other studies [119,121,136]. According to the summary of the correlation of the optimum efficiency of power dissipation in the DRX region with normalized stacking fault energy in various metals, the optimum efficiency for Ti was estimated to be in the range of 40%–50% [137].

3.3 Rheological instability parameter and typical instability phenomenon

According to Eq. (7), the calculated rheological

instability parameters are usually closely related to wedge-shaped cracks, voids, and adiabatic shear bands in the deformation process. As one of the difficult deformation materials, β titanium alloy should avoid plastic flow instability in the process of hot working. According to Table 2 and Fig. 11(b), the instability region is mainly concentrated in the high strain rate region. On the one hand, the increase in deformation temperature caused by deformation heating enhances the dynamic softening effect and offsets the stabilizing effect of work hardening on deformation. On the other hand, it is caused by the increase in deformation resistance at high strain rates.

Figure 13 shows the typical instability phenomena of several β titanium alloys. Figure 13(a) shows 45° shear cracking of Ti-5553 alloy at a temperature of 700 °C and a strain rate of 10 s⁻¹. Figures 13(b, c) show longitudinal cracking. Unlike a 45° shear fracture, the free-surface longitudinal cracking is mainly caused by secondary tensile stresses. These stresses mainly come from the upsetting of the specimen during uniaxial hot compression, and they are applied perpendicularly to the compression direction [138]. Figures 13(d, e)

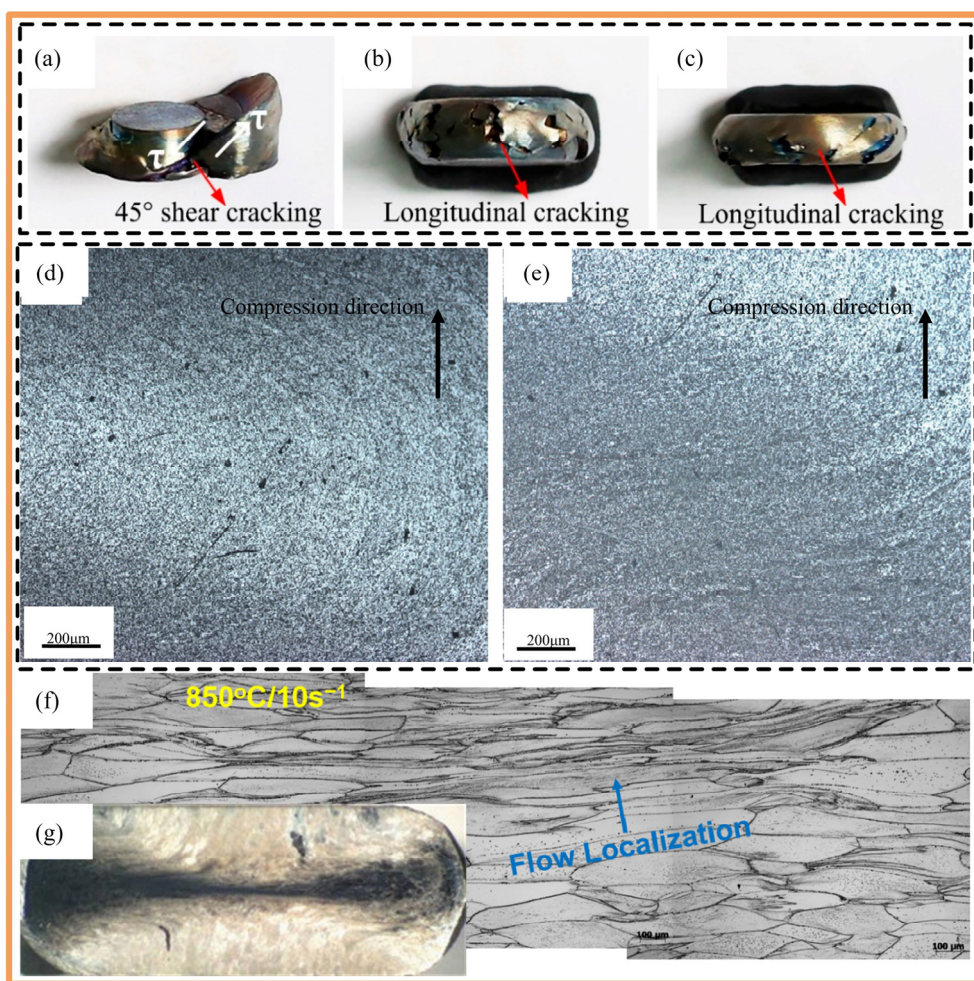


Fig. 13 Typical instability phenomena: (a) Shear cracking; (b, c) Longitudinal cracking [138]; (d, e) Adiabatic shear bands [42]; (f, g) Flow localization [128,139]

show the adiabatic shear bands of Ti-55511 alloy along approximately 45° to the compression axis. Similar phenomena also appeared in other studies [94,127,140,141]. In addition, HUANG et al [142] found that the DRX phenomenon also exists in the center of the adiabatic shear band, and the LAGB mainly exists in the β phase. Figures 13(f, g) show the flow localization of titanium alloy at a high strain rate. Flow localization is formed due to the shear force caused by the increase in adiabatic temperature rise and low thermal conductivity in the process of high strain rate deformation [22,52,143]. This phenomenon is also very common in beta titanium alloys [53,60,91,109,123–126,129,130,144,145]. LIANG et al [134] also found grain coarsening at high temperatures and low strain rates, which was caused by element diffusion and grain rotation. In addition, the instability region was concentrated in

the strain rate range of $0.001\text{--}1\text{ s}^{-1}$ at low strain in Ti-5Al-4Zr-8Mo-7V alloy. When the strain was further increased to 1.0, the instability region occurred in the strain rate range of $0.0018\text{--}0.018\text{ s}^{-1}$ and temperature range of $815\text{--}835\text{ }^{\circ}\text{C}$, and the area of the instability region decreased at a high strain rate. There are a few studies [139,146,147] about this phenomenon. Considering that hot deformation is a continuous process, the instability region at low strain may disappear with increasing strain. Therefore, it is necessary to regulate and control the instability in hot deformation more accurately, which is also one of the research difficulties.

4 Summary and prospect

The microstructure evolution, deformation mechanism and hot processing map of high strength β titanium alloys during high temperature plastic

deformation in recent years are reviewed in this paper. Many researchers have conducted in-depth studies on the microstructure evolution behavior and deformation mechanism during the deformation process through various characterization methods as a critical factor affecting mechanical properties and flow softening. Moreover, the deformation mechanism can be revealed, and the defects in the deformation process can be predicted by combining the hot processing map, which is of great significance to reduce production costs and improve production efficiency. Generally, there still exist some problems in the hot deformation behavior of high strength β titanium alloys based on hot compression experiments.

(1) There is a lack of research on the influence of the interaction of different softening mechanisms on flow stress. During the hot deformation of titanium alloy, softening mechanisms such as dynamic recovery, dynamic recrystallization, and dynamic phase transformation will appear. Most current research explores a particular softening mechanism through thermodynamic parameters and microstructure characterization under a single deformation condition. However, it is difficult to capture the contribution of different softening mechanisms to the flow stress. Furthermore, it is challenging to clarify the interaction between various softening mechanisms and precisely control the softening behavior.

(2) There is a lack of research on the formation mechanism and evolution law of instability phenomena such as adiabatic shear band and flow localization. High strength β titanium alloy, as a difficult-to-deform material, is prone to obvious deformation defects such as adiabatic shear bands and flow localization during hot deformation. Most of the current research focuses on the macro and micro characterization of samples at the end of deformation to determine the instability phenomenon and analyze the causes from the perspective of thermodynamics. However, the critical conditions for the occurrence of instability and its microstructural heredity lack corresponding theoretical explanations. Meanwhile, the plastic instability regions in the hot processing map will also appear or disappear along with strain changes, and the intrinsic mechanism lacks corresponding research. Based on the problems mentioned above, this article proposes the following important

development directions.

Establishment of an evaluation system for accurately predicting the hot deformation instability of high strength β titanium alloys. At present, hot processing maps can predict the hot deformation instability of high strength β titanium alloys. However, the dynamic material model is based on the plastic flow and does not consider the microstructure evolution mechanism during the deformation process, which leads to the deviation between the optimization of machining parameters and the prediction of instability. Therefore, it is necessary to combine the dynamic material model and the irreversible thermodynamic extreme value theory, couple with the microstructure evolution mechanism, and optimize the calculation method of the energy dissipation rate and flow instability criterion. In this way, a higher precision plastic deformation hot processing map is established, which provides a more effective solution for predicting the formation of defects during hot processing.

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References

- [1] YUMAK N, ASLANTAŞ K. A review on heat treatment efficiency in metastable β titanium alloys: The role of treatment process and parameters [J]. *Journal of Materials Research and Technology*, 2020, 9: 15360–15380.
- [2] LONG M, RACK H J. Titanium alloys in total joint replacement—A materials science perspective [J]. *Biomaterials*, 1998, 19: 1621–1639.
- [3] WANG K. The use of titanium for medical applications in the USA [J]. *Materials Science and Engineering A*, 1996, 213: 134–137.
- [4] DONACHIE M J Jr. *Titanium: A technical guide* [M]. 2nd ed. Materials Park, Ohio: ASM International, 2000.
- [5] BANIA P J. Beta titanium alloys and their role in the titanium industry [J]. *JOM*, 1994, 46: 16–19.
- [6] KOLLI R, DEVARAJ A. A review of metastable beta

- titanium alloys [J]. *Metals*, 2018, 8: 506.
- [7] EYLLON D, VASSEL A, COMBRES Y, BOYER R R, BANIA P J, SCHUTZ R W. Issues in the development of beta titanium alloys [J]. *JOM*, 1994, 46: 14–15.
 - [8] BOYER R R, BRIGGS R D. The use of β titanium alloys in the aerospace industry [J]. *Journal of Materials Engineering and Performance*, 2005, 14: 681–685.
 - [9] MARKOVSKY P E, BONDARCHUK V I. Influence of strain rate, microstructure and chemical and phase composition on mechanical behavior of different titanium alloys [J]. *Journal of Materials Engineering and Performance*, 2017, 26: 3431–3449.
 - [10] SCHMIDT P, EL-CHAIKH A, CHRIST H J. Effect of duplex aging on the initiation and propagation of fatigue cracks in the solute-rich metastable β titanium alloy Ti 38-644 [J]. *Metallurgical and Materials Transactions A*, 2011, 42: 2652–2667.
 - [11] LI Cheng-lin, MI Xu-jun, YE Wen-jun, HUI Song-xiao, YU Yang, WANG Wei-qi. A study on the microstructures and tensile properties of new beta high strength titanium alloy [J]. *Journal of Alloys and Compounds*, 2013, 550: 23–30.
 - [12] COTTON J D, BRIGGS R D, BOYER R R, TAMIRISAKANDALA S, RUSSO P, SHCHETNIKOV N, FANNING J C. State of the art in beta titanium alloys for airframe applications [J]. *JOM*, 2015, 67: 1281–1303.
 - [13] CHOU Y K, TSAY L W, CHEN C. Effects of aging treatments on the mechanical behavior of Ti–15V–3Cr–3Sn–3Al alloy [J]. *Journal of Materials Engineering and Performance*, 2015, 24: 3365–3372.
 - [14] SUN Meng, LI Dong, GUO Yang-hua, WANG Ying, DONG Yue-cheng, DAN Zhen-hua, CHANG Hui. The effect of heat treatment on the microstructure and mechanical properties of the novel low-cost Ti–3Al–5Mo–4Cr–2Zr–1Fe alloy [J]. *Materials (Basel)*, 2020, 13: 3798.
 - [15] LÜ Zhi-dan, ZHANG Chang-jiang, DU Zhao-xin, HAN Jian-chao, ZHANG Shu-zhi, YANG Fei, CHEN Yu-yong. Relationship between microstructure and tensile properties on a near- β titanium alloy after multidirectional forging and heat treatment [J]. *Rare Metals*, 2019, 38: 336–342.
 - [16] FAN Jiang-kun, LI Jin-shan, KOU Hong-chao, HUA Ke, TANG Bin, ZHANG Yu-dong. Influence of solution treatment on microstructure and mechanical properties of a near β titanium alloy Ti-7333 [J]. *Materials & Design*, 2015, 83: 499–507.
 - [17] YU Ming, LIN Chen-guang, LI Feng, CUI Xue-fei, LUO Zhen. Effect of heat treatment on microstructure and properties of Ti–5Mo–5V–2Cr–3Al alloy [J]. *Chinese Journal of Rare Metals*, 2009, 33: 790–794. (in Chinese)
 - [18] LU Jin-wen, ZHAO Yong-qing, GE Peng, ZHANG Yu-sheng, NIU Hong-zhi, ZHANG Wei, ZHANG Ping-xiang. Precipitation behavior and tensile properties of new high strength beta titanium alloy Ti-1300 [J]. *Journal of Alloys and Compounds*, 2015, 637: 1–4.
 - [19] RAN Chun, SHENG Ze-min, CHEN Peng-wan, ZHANG Wang-feng, CHEN Qi. Effect of microstructure on the mechanical properties of Ti–5Al–5Mo–5V–1Cr–1Fe alloy [J]. *Materials Science and Engineering A*, 2020, 773: 138728.
 - [20] ZHU Wen-guang, LEI Jia, TAN Chang-sheng, SUN Qiao-yan, CHEN Wei, XIAO Lin, SUN Jun. A novel high-strength β -Ti alloy with hierarchical distribution of α -phase: The superior combination of strength and ductility [J]. *Materials & Design*, 2019, 168: 107640.
 - [21] AN Zhen, LI J S, FENG Yong, LIU X H, DU Y X. Characterization of hot deformation behavior of a new near- β titanium alloy: Ti555211 [J]. *High Temperature Materials and Processes*, 2015, 35: 913–928.
 - [22] WANG Zhe, WANG Xin-nan, ZHU Zhi-shou. Characterization of high-temperature deformation behavior and processing map of TB17 titanium alloy [J]. *Journal of Alloys and Compounds*, 2017, 692: 149–154.
 - [23] LI Hui-min, LI Miao-quan, LUO Jiao, WANG Ke. Microstructure and mechanical properties of heat-treated Ti–5Al–2Sn–2Zr–4Mo–4Cr [J]. *Transactions of Nonferrous Metals Society of China*, 2015, 25: 2893–2900.
 - [24] WU Chuan, ZHAN Mei. Microstructural evolution, mechanical properties and fracture toughness of near β titanium alloy during different solution plus aging heat treatments [J]. *Journal of Alloys and Compounds*, 2019, 805: 1144–1160.
 - [25] ALLUAIBI M H I, COJOCARU E M, RUSEA A, ȘERBAN N, COMAN G, COJOCARU V D. Microstructure and mechanical properties evolution during solution and ageing treatment for a hot deformed, above β -transus, Ti-6246 alloy [J]. *Metals*, 2020, 10: 1114.
 - [26] SHI Shuang-xi, LIU Xiu-sheng, ZHANG Xiao-yong, ZHOU Ke-chao. Comparison of flow behaviors of near beta Ti-55511 alloy during hot compression based on SCA and BPANN models [J]. *Transactions of Nonferrous Metals Society of China*, 2021, 31: 1665–1679.
 - [27] XIA Yu-feng, JIANG Wei, CHENG Qian, JIANG Lai, JIN Li. Hot deformation behavior of Ti–6Al–4V–0.1Ru alloy during isothermal compression [J]. *Transactions of Nonferrous Metals Society of China*, 2020, 30: 134–146.
 - [28] WANG Ke, WU Ming-yu, REN Zhao, ZHANG Yu, XIN Ren-long, LIU Qing. Static globularization and grain morphology evolution of α and β phases during annealing of hot-rolled TC21 titanium alloy [J]. *Transactions of Nonferrous Metals Society of China*, 2021, 31: 2664–2676.
 - [29] FAN X G, ZHANG Y, GAO P F, LEI Z N, ZHAN M. Deformation behavior and microstructure evolution during hot working of a coarse-grained Ti–5Al–5Mo–5V–3Cr–1Zr titanium alloy in beta phase field [J]. *Materials Science and Engineering A*, 2017, 694: 24–32.
 - [30] KOLLI R P, JOOST W J, ANKEM S. Phase stability and stress-induced transformations in beta titanium alloys [J]. *JOM*, 2015, 67: 1273–1280.
 - [31] SARMA J, KUMAR R, SAHOO A K, PANDA A. Enhancement of material properties of titanium alloys through heat treatment process: A brief review [J]. *Materials Today: Proceedings*, 2020, 23: 561–564.
 - [32] LI Chang-min, HUANG Liang, ZHAO Ming-jie, GUO Shi-qi, LI Jian-jun. Hot deformation behavior and mechanism of a new metastable β titanium alloy Ti–6Cr–5Mo–5V–4Al in single phase region [J]. *Materials Science and Engineering A*, 2021, 814: 141231.
 - [33] LIN Y C, HUANG J, HE D G, ZHANG X Y, WU Q, WANG L H, CHEN C, ZHOU K C. Phase transformation and

- dynamic recrystallization behaviors in a Ti55511 titanium alloy during hot compression [J]. *Journal of Alloys and Compounds*, 2019, 795: 471–482.
- [34] LI L, LI M Q, LUO J. Mechanism in the β phase evolution during hot deformation of Ti–5Al–2Sn–2Zr–4Mo–4Cr with a transformed microstructure [J]. *Acta Materialia*, 2015, 94: 36–45.
- [35] ZHAO Qin-yang, YANG Fei, TORRENS R, BOLZONI L. Comparison of hot deformation behaviour and microstructural evolution for Ti–5Al–5V–5Mo–3Cr alloys prepared by powder metallurgy and ingot metallurgy approaches [J]. *Materials & Design*, 2019, 169: 107682.
- [36] FURUHARA T, POORGANJI B, ABE H, MAKI T. Dynamic recovery and recrystallization in titanium alloys by hot deformation [J]. *JOM*, 2007, 59: 64–67.
- [37] BAO R Q, HUANG Xu, CAO C X. Deformation behavior and mechanisms of Ti-1023 alloy [J]. *Transactions of Nonferrous Metals Society of China*, 2006, 16: 274–280.
- [38] FAN J K, KOU H C, LAI M J, TANG B, CHANG H, LI J S. Hot deformation mechanism and microstructure evolution of a new near β titanium alloy [J]. *Materials Science and Engineering A*, 2013, 584: 121–132.
- [39] WANG Q W, LIN Y C, JIANG Y Q, LIU X G, ZHANG X Y, CHEN D D, CHEN C, ZHOU K C. Precipitation behavior of a β -quenched Ti–5Al–5Mo–5V–1Cr–1Fe alloy during high-temperature compression [J]. *Materials Characterization*, 2019, 151: 358–367.
- [40] QU F S, ZHOU Y H, ZHANG L Y, WANG Z H, ZHOU J. Research on hot deformation behavior of Ti–5Al–5Mo–5V–1Cr–1Fe alloy [J]. *Materials & Design*, 2015, 69: 153–162.
- [41] LUO J, LIU S F, LI M Q. Quantitative analysis of microstructure and deformation mechanisms during isothermal compression of Ti–5Al–5Mo–5V–1Cr–1Fe alloy [J]. *Materials Characterization*, 2015, 108: 115–123.
- [42] NIE Xi-an, HU Zhan, LIU Hui-qun, YI Dan-qing, CHEN Ti-ying, WANG Bing-feng, GAO Qi, WANG Ding-chun. High temperature deformation and creep behavior of Ti–5Al–5Mo–5V–1Fe–1Cr alloy [J]. *Materials Science and Engineering A*, 2014, 613: 306–316.
- [43] MATSUMOTO H, KITAMURA M, LI Y P, KOIZUMI Y, CHIBA A. Hot forging characteristic of Ti–5Al–5V–5Mo–3Cr alloy with single metastable β microstructure [J]. *Materials Science and Engineering A*, 2014, 611: 337–344.
- [44] HU Zhan, ZHOU X Y, NIE X A, ZHAO S Y, LIU H Q, YI D Q, ZHANG X Y. Finer subgrain microstructure induced by multi-pass compression in $\alpha+\beta$ phase region in a near- β Ti–5Al–5Mo–5V–1Cr–1Fe alloy [J]. *Journal of Alloys and Compounds*, 2019, 788: 136–147.
- [45] CALLEGARI B, OLIVEIRA J P, COELHO R S, BRITO P P, SCHELL N, SOLDERA F A, MÜCKLICH F, SADIK M I, GARCÍA J L, PINTO H C. New insights into the microstructural evolution of Ti–5Al–5Mo–5V–3Cr alloy during hot working [J]. *Materials Characterization*, 2020, 162: 110180.
- [46] XU Xin, DONG Li-min, BA Hong-bo, ZHANG Zhi-qiang, YANG Rui. Hot deformation behavior and microstructural evolution of beta C titanium alloy in β phase field [J]. *Transactions of Nonferrous Metals Society of China*, 2016, 26: 2874–2882.
- [47] LIANG Hou-quan, GUO Hong-zhen, NING Yong-quan, PENG Xiao-na, QIN Chun, SHI Zhi-feng, NAN Yang. Dynamic recrystallization behavior of Ti–5Al–5Mo–5V–1Cr–1Fe alloy [J]. *Materials & Design*, 2014, 63: 798–804.
- [48] MECKING H, KOCKS U F. Kinetics of flow and strain-hardening [J]. *Acta Metallurgica*, 1981, 29: 1865–1875.
- [49] MOMENI A, EBRAHIMI G R, JAHAZI M, BOCHER P. Microstructure evolution at the onset of discontinuous dynamic recrystallization: A physics-based model of subgrain critical size [J]. *Journal of Alloys and Compounds*, 2014, 587: 199–210.
- [50] LI Miao-quan, PAN Hong-si, LIN Ying-ying, LUO Jiao. High temperature deformation behavior of near alpha Ti–5.6Al–4.8Sn–2.0Zr alloy [J]. *Journal of Materials Processing Technology*, 2007, 183: 71–76.
- [51] MENG L, KITASHIMA T, TSUCHIYAMA T, WATANABE M. Effect of α precipitation on β texture evolution during β -processed forging in a near- β titanium alloy [J]. *Materials Science and Engineering A*, 2020, 771: 138640.
- [52] LI L, LI M Q, LUO J. Flow softening mechanism of Ti–5Al–2Sn–2Zr–4Mo–4Cr with different initial microstructures at elevated temperature deformation [J]. *Materials Science and Engineering A*, 2015, 628: 11–20.
- [53] LIU S F, LI M Q, LUO J, YANG Z. Deformation behavior in the isothermal compression of Ti–5Al–5Mo–5V–1Cr–1Fe alloy [J]. *Materials Science and Engineering A*, 2014, 589: 15–22.
- [54] POLETTI C, GERMAIN L, WARCHOMICKA F, DIKOVITS M, MITSCHKE S. Unified description of the softening behavior of beta-metastable and alpha + beta titanium alloys during hot deformation [J]. *Materials Science and Engineering A*, 2016, 651: 280–290.
- [55] PHILIPPART I, RACK H J. High temperature, high strain deformation behavior of Ti–6.8Mo–4.5Fe–1.5Al [J]. *Materials Science and Engineering A*, 1998, 254: 253–267.
- [56] YAN C K, FENG A H, QU S J, CAO G J, SUN J L, SHEN J, CHEN D L. Dynamic recrystallization of titanium: Effect of pre-activated twinning at cryogenic temperature [J]. *Acta Materialia*, 2018, 154: 311–324.
- [57] WANG X, WANG L, LUO L S, LIU X D, TANG Y C, LI X Z, CHEN R R, SU Y Q, GUO J J, FU H Z. Hot deformation behavior and dynamic recrystallization of melt hydrogenated Ti–6Al–4V alloy [J]. *Journal of Alloys and Compounds*, 2017, 728: 709–718.
- [58] ZHOU Hai-tao, KONG Fan-tao, WU Kai, WANG Xiao-peng, CHEN Yu-yong. Hot pack rolling nearly lamellar Ti–44Al–8Nb–(W, B, Y) alloy with different rolling reductions: Lamellar colonies evolution and tensile properties [J]. *Materials & Design*, 2017, 121: 202–212.
- [59] TAN K, LI J, GUAN Z J, YANG J B, SHU J X. The identification of dynamic recrystallization and constitutive modeling during hot deformation of Ti55511 titanium alloy [J]. *Materials & Design*, 2015, 84: 204–211.
- [60] ZHAO Q Y, YANG F, TORRENS R, BOLZONI L. Evaluation of the hot workability and deformation mechanisms for a metastable beta titanium alloy prepared from powder [J]. *Materials Characterization*, 2019, 149: 226–238.

- [61] ABBASI S M, MOMENI A, LIN Y C, JAFARIAN H R. Dynamic softening mechanism in Ti–13V–11Cr–3Al beta Ti alloy during hot compressive deformation [J]. *Materials Science and Engineering A*, 2016, 665: 154–160.
- [62] OUYANG D L, FU M W, LU S Q. Study on the dynamic recrystallization behavior of Ti-alloy Ti–10V–2Fe–3V in β processing via experiment and simulation [J]. *Materials Science and Engineering A*, 2014, 619: 26–34.
- [63] SUN Yong-gang, ZHANG Chang-jiang, FENG Hong, ZHANG Shu-zhi, HAN Jian-chao, ZHANG Wang-gang, ZHAO Er-tuan, WANG Hong-wei. Dynamic recrystallization mechanism and improved mechanical properties of a near α high temperature titanium alloy processed by severe plastic deformation [J]. *Materials Characterization*, 2020, 163: 110281.
- [64] LIN Y C, PANG G D, JIANG Y Q, LIU X G, ZHANG X Y, CHEN C, ZHOU K C. Hot compressive deformation behavior and microstructure evolution of a Ti-55511 alloy with basket-weave microstructures [J]. *Vacuum*, 2019, 169: 108878.
- [65] SUN J Z, LI M Q, LI H. Initial flow softening and restoration mechanisms of isothermally compressed Ti–5Al–2Sn–2Zr–4Mo–4Cr with basketweave microstructure [J]. *Materials Science and Engineering A*, 2017, 697: 132–140.
- [66] ABBASI S M, MORAKKABATI M, SHEIKHALI A H, MOMENI A. Hot deformation behavior of beta titanium Ti–13V–11Cr–3Al alloy [J]. *Metallurgical and Materials Transactions A*, 2014, 45: 5201–5211.
- [67] ZHAO Juan, ZHONG Jie, YAN Fei, CHAI Fang, DARGUSCH M. Deformation behaviour and mechanisms during hot compression at supertransus temperatures in Ti–10V–2Fe–3Al [J]. *Journal of Alloys and Compounds*, 2017, 710: 616–627.
- [68] ZHAO Y H, GE P, ZHAO Y Q, YANG G J, WENG J H. Hot deformation behavior of Ti-1300 alloy [J]. *Rare Metal Materials and Engineering*, 2009, 38: 46–49. (in Chinese)
- [69] MEHTONEN S V, PALMIERE E J, MISRA R D K, KARJALAINEN L P, PORTER D A. Dynamic restoration mechanisms in a Ti–Nb stabilized ferritic stainless steel during hot deformation [J]. *Materials Science and Engineering A*, 2014, 601: 7–19.
- [70] CHAO Q, HODGSON P D, BELADI H. Ultrafine grain formation in a Ti–6Al–4V alloy by thermomechanical processing of a martensitic microstructure [J]. *Metallurgical and Materials Transactions A*, 2014, 45: 2659–2671.
- [71] SUN J Z, LI M Q, LI H. Interaction effect between alpha and beta phases based on dynamic recrystallization of isothermally compressed Ti–5Al–2Sn–2Zr–4Mo–4Cr with basketweave microstructure [J]. *Journal of Alloys and Compounds*, 2017, 692: 403–412.
- [72] MOMENI A, ABBASI S M, MORAKABATI M, GHAZI MIRSAED S M. Flow softening behavior of Ti–13V–11Cr–3Al beta Ti alloy in double-hit hot compression tests [J]. *Journal of Materials Research*, 2016, 31: 3900–3906.
- [73] XU Jian-wei, ZENG Wei-dong, ZHOU Da-di, MA Hao-yuan, HE Sheng-tong, CHEN Wei. Analysis of flow softening during hot deformation of Ti-17 alloy with the lamellar structure [J]. *Journal of Alloys and Compounds*, 2018, 767: 285–292.
- [74] MOMENI A, ABBASI S M, SADEGHPOUR S. A comparative study on the hot deformation behavior of Ti₅Al₅Mo₅V₃Cr and newly developed Ti₄Al₇Mo₃V₃Cr alloys [J]. *Vacuum*, 2019, 161: 410–418.
- [75] LI L, LUO J, YAN J J, LI M Q. Dynamic globularization and restoration mechanism of Ti–5Al–2Sn–2Zr–4Mo–4Cr alloy during isothermal compression [J]. *Journal of Alloys and Compounds*, 2015, 622: 174–183.
- [76] YADA H, LI C M, YAMAGATA H. Dynamic $\gamma \rightarrow \alpha$ transformation during hot deformation in iron–nickel–carbon alloys [J]. *ISIJ International*, 2000, 40: 200–206.
- [77] ARANAS C Jr, FOUL A, GUO B Q, FALL A, JAHAZI M, JONAS J J. Determination of the critical stress for the initiation of dynamic transformation in commercially pure titanium [J]. *Scripta Materialia*, 2017, 133: 83–85.
- [78] JI Xian-kun, GUO Bao-qi, JIANG Fu-lin, YU Hong, FU Ding-fa, TENG Jie, ZHANG Hui, JONAS J J. Accelerated flow softening and dynamic transformation of Ti–6Al–4V alloy in two-phase region during hot deformation via coarsening α grain [J]. *Journal of Materials Science & Technology*, 2020, 36: 160–166.
- [79] GUO B Q, SEMIATIN S L, JONAS J J. Dynamic transformation during the high temperature deformation of two-phase titanium alloys [J]. *Materials Science and Engineering A*, 2019, 761: 138047.
- [80] KOIKE J, SHIMOYAMA Y, OHNUMA I, OKAMURA T, KAINUMA R, ISHIDA K, MARUYAMA K. Stress-induced phase transformation during superplastic deformation in two-phase Ti–Al–Fe alloy [J]. *Acta Materialia*, 2000, 48: 2059–2069.
- [81] HUA Ke, ZHANG Yu-dong, GAN Wei-min, KOU Hong-chao, BEAUSIR B, LI Jin-shan, ESLING C. Hot deformation behavior originated from dislocation activity and β to α phase transformation in a metastable β titanium alloy [J]. *International Journal of Plasticity*, 2019, 119: 200–214.
- [82] LI Kai, YANG Ping. Strain-induced α -to- β phase transformation during hot compression in Ti–5Al–5Mo–5V–1Cr–1Fe alloy [J]. *Transactions of Nonferrous Metals Society of China*, 2019, 29: 296–304.
- [83] AN Zhen, LI Jin-shan, FENG Yong. Microstructure evolution of a new near- β titanium alloy: Ti555211 during high-temperature deformation [J]. *Rare Metals*, 2015, 34: 757–763.
- [84] WU Chuan, HUANG Liang, LI Chang-min. Experimental investigation on dynamic phase transformation and texture evolution of Ti55531 high strength titanium alloy during hot compression in the $\alpha + \beta$ region [J]. *Materials Science and Engineering A*, 2020, 773: 138851.
- [85] HUA Ke, XUE Xiang-yi, KOU Hong-chao, FAN Jiang-kun, TANG Bin, LI Jin-shan. Characterization of hot deformation microstructure of a near beta titanium alloy Ti-5553 [J]. *Journal of Alloys and Compounds*, 2014, 615: 531–537.
- [86] HUA Ke, LI Jin-shan, KOU Hong-chao, FAN Jiang-kun, SUN Miao, TANG Bin. Phase precipitation behavior during isothermal deformation in β -quenched near beta titanium alloy Ti-7333 [J]. *Journal of Alloys and Compounds*, 2016, 671: 381–388.

- [87] ZHU Z S, GU J L, CHEN N P. Variant selection in $\alpha \rightarrow \beta \rightarrow \alpha$ phase transformation of cold rolled titanium sheet [J]. Scripta Materialia, 1996, 34: 1281–1286.
- [88] PANG G D, LIN Y C, JIANG Y Q, ZHANG X Y, LIU X G, XIAO Y W, ZHOU K C. Precipitation behaviors and orientation evolution mechanisms of α phases in Ti-55511 titanium alloy during heat treatment and subsequent hot deformation [J]. Materials Characterization, 2020, 167: 110471.
- [89] MIRONOV S, MURZINOVA M, ZHEREBTSOV S, SALISHCHEV G A, SEMIATIN S L. Microstructure evolution during warm working of Ti-6Al-4V with a colony- α microstructure [J]. Acta Materialia, 2009, 57: 2470–2481.
- [90] HE L J, DEGHAN-MANSHADI A, DIPPENAAR R J. The evolution of microstructure of Ti-6Al-4V alloy during concurrent hot deformation and phase transformation [J]. Materials Science and Engineering A, 2012, 549: 163–167.
- [91] LI Lian, LI Miao-quan. Constitutive model and optimal processing parameters of TC17 alloy with a transformed microstructure via kinetic analysis and processing maps [J]. Materials Science and Engineering A, 2017, 698: 302–312.
- [92] KLIMOVA M, ZHEREBTSOV S, SALISHCHEV G, SEMIATIN S L. Influence of deformation on the Burgers orientation relationship between the α and β phases in Ti-5Al-5Mo-5V-1Cr-1Fe [J]. Materials Science and Engineering A, 2015, 645: 292–297.
- [93] WANG Kai-xuan, ZENG Wei-dong, ZHAO Yong-qing, LAI Yun-jin, ZHOU Yi-gang. Dynamic globularization kinetics during hot working of Ti-17 alloy with initial lamellar microstructure [J]. Materials Science and Engineering A, 2010, 527: 2559–2566.
- [94] WANG Kai-xuan, ZENG Wei-dong, ZHAO Yong-qing, LAI Yun-jin, ZHOU Yi-gang. Hot working of Ti-17 titanium alloy with lamellar starting structure using 3-D processing maps [J]. Journal of Materials Science, 2010, 45: 5883–5891.
- [95] LI Chao, ZHANG Xiao-yong, ZHOU Ke-chao, PENG Chao-qun. Relationship between lamellar α evolution and flow behavior during isothermal deformation of Ti-5Al-5Mo-5V-1Cr-1Fe near β titanium alloy [J]. Materials Science and Engineering A, 2012, 558: 668–674.
- [96] LI Shao-jun, LV Ya-ping, ZHANG Xiao-yong, CHEN Chao, ZHOU Ke-chao. Twinning and its effect on the evolution of acicular α in Ti-5Al-5Mo-5V-3Cr-1Zr [J]. Materials Science and Technology, 2019, 35: 839–845.
- [97] BUZOLIN R H, LASNIK M, KRUMPHALS A, POLETTI M C. Hot deformation and dynamic α -globularization of a Ti-17 alloy: Consistent physical model [J]. Materials & Design, 2021, 197: 109266.
- [98] JONES N G, DASHWOOD R J, DYE D, JACKSON M. Thermomechanical processing of Ti-5Al-5Mo-5V-3Cr [J]. Materials Science and Engineering A, 2008, 490: 369–377.
- [99] JACKSON M, DASHWOOD R, FLOWER H, CHRISTODOULOU L. The microstructural evolution of near beta alloy Ti-10V-2Fe-3Al during subtransus forging [J]. Metallurgical and Materials Transactions A, 2005, 36: 1317–1327.
- [100] SONG Hong-wu, ZHANG Shi-hong, CHENG Ming. Dynamic globularization kinetics during hot working of a two phase titanium alloy with a colony alpha microstructure [J]. Journal of Alloys and Compounds, 2009, 480: 922–927.
- [101] CHEN S H, ZHANG M C, JIA M L, LI W. Investigations on microstructure and grain orientation uniformity of Ti-17 titanium alloy under different hot deformation modes [J]. Journal of Materials Processing Technology, 2018, 252: 148–158.
- [102] JONES N G, DASHWOOD R J, DYE D, JACKSON M. The flow behavior and microstructural evolution of Ti-5Al-5Mo-5V-3Cr during subtransus isothermal forging [J]. Metallurgical and Materials Transactions A, 2009, 40: 1944–1954.
- [103] JACKSON M, JONES N G, DYE D, DASHWOOD R J. Effect of initial microstructure on plastic flow behaviour during isothermal forging of Ti-10V-2Fe-3Al [J]. Materials Science and Engineering A, 2009, 501: 248–254.
- [104] LI Shao-jun, LV Ya-ping, ZHANG Xiao-yong, ZHOU Ke-chao. Relationship between flow behavior and microstructure evolution during isothermal compression of near β titanium alloy Ti-55531 with acicular starting microstructure [J]. Metals, 2018, 8: 467.
- [105] LI Chao, ZHANG Xiao-yong, LI Zhi-you, ZHOU Ke-chao. Hot deformation of Ti-5Al-5Mo-5V-1Cr-1Fe near β titanium alloys containing thin and thick lamellar α phase [J]. Materials Science and Engineering A, 2013, 573: 75–83.
- [106] CHEN Ke, LUO Jiao, HAN Wen-chao, LI Miao-quan. Formation and evolution of new α grain boundary and its influence on globularization of α lamellae in TC17 alloy [J]. Journal of Alloys and Compounds, 2020, 848: 156141.
- [107] LUO J, YE P, HAN W C, LI M Q. Microstructure evolution and its effect on flow stress of TC17 alloy during deformation in $\alpha+\beta$ two-phase region [J]. Transactions of Nonferrous Metals Society of China, 2019, 29: 1430–1438.
- [108] MU Z, LI H, LI M Q. The microstructure evolution in the isothermal compression of Ti-17 alloy [J]. Materials Science and Engineering A, 2013, 582: 108–116.
- [109] FAN J K, KOU H C, LAI M J, TANG B, CHANG H, LI J S. Characterization of hot deformation behavior of a new near beta titanium alloy: Ti-7333 [J]. Materials & Design, 2013, 49: 945–952.
- [110] QUAN Guo-zheng, PU Shi-ao, WEN Hai-rong, ZOU Zhen-yu, ZHOU Jie. Quantitative analysis of dynamic softening behaviors induced by dynamic recrystallization for Ti-10V-2Fe-2Al alloy [J]. High Temperature Materials and Processes, 2015, 34: 549–561.
- [111] SESHACHARYULU T, MEDEIROS S C, MORGAN J T, MALAS J C, FRAZIER W G, PRASAD Y V R K. Hot deformation mechanisms in ELI grade Ti-6Al-4V [J]. Scripta Materialia, 1999, 41: 283–288.
- [112] LIN Y C, XIAO Y W, JIANG Y Q, PANG G D, LI H B, ZHANG X Y, ZHOU K C. Spheroidization and dynamic recrystallization mechanisms of Ti-55511 alloy with bimodal microstructures during hot compression in $\alpha+\beta$ region [J]. Materials Science and Engineering A, 2020, 782: 139282.
- [113] LONG Shuai, XIA Yu-feng, WANG Peng, ZHOU Yu-ting, GONG-YE Fan-jiao, ZHOU Jie, ZHANG Jian-sheng, CUI Ming-liang. Constitutive modelling, dynamic globularization behavior and processing map for Ti-6Cr-5Mo-5V-4Al

- alloy during hot deformation [J]. *Journal of Alloys and Compounds*, 2019, 796: 65–76.
- [114] CHEN Jia-hao, LI Jin-shan, TANG Bin, CHEN Yi, KOU Hong-chao. Microstructure and texture evolution of a near β titanium alloy Ti-7333 during continuous cooling hot deformation [J]. *Progress in Natural Science: Materials International*, 2019, 29: 50–56.
- [115] LV Ya-ping, LI Shao-jun, ZHANG Xiao-yong, CHEN Chao, ZHOU Ke-chao. Effects of strain rate on grain boundary evolution in Ti-55531 alloy [J]. *Materials Science and Technology*, 2018, 34: 1880–1886.
- [116] LI Kai, YANG Ping. The formation of strong {100} texture by dynamic strain-induced boundary migration in hot compressed Ti-5Al-5Mo-5V-1Cr-1Fe alloy [J]. *Metals*, 2017, 7: 412.
- [117] PRASAD Y, SESHACHARYULU T. Modelling of hot deformation for microstructural control [J]. *International Materials Reviews*, 1998, 43: 243–258.
- [118] MURTY S V S, RAO B N. Ziegler's criterion on the instability regions in processing maps [J]. *Journal of Materials Science Letters*, 1998, 17: 1203–1205.
- [119] QUAN Guo-zheng, LV Wen-quan, LIANG Jian-ting, PU Shi-ao, LUO Gui-chang, LIU Qing. Evaluation of the hot workability corresponding to complex deformation mechanism evolution for Ti-10V-2Fe-3Al alloy in a wide condition range [J]. *Journal of Materials Processing Technology*, 2015, 221: 66–79.
- [120] CHANG Li-li, ZHENG Li-wei. Isothermal compression behavior and constitutive modeling of Ti-5Al-5Mo-5V-1Cr-1Fe alloy [J]. *Transactions of Nonferrous Metals Society of China*, 2018, 28: 1114–1122.
- [121] DIKOVITS M, POLETTI C, WARCHOMICKA F. Deformation mechanisms in the near- β titanium alloy Ti-55531 [J]. *Metallurgical and Materials Transactions A*, 2014, 45: 1586–1596.
- [122] WU Chuan, HUANG Liang. Hot deformation and dynamic recrystallization of a near-beta titanium alloy in the β single phase region [J]. *Vacuum*, 2018, 156: 384–401.
- [123] AN Zhen, LI Jin-shan, FENG Yong. Optimization of thermal processing parameters of Ti555211 alloy using processing maps based on Murty criterion [J]. *Rare Metals*, 2016, 35: 154–161.
- [124] WU Jie, LÜ Zhi-dan, ZHANG Chang-jiang, HAN Jian-chao, ZHANG Hong-zhou, ZHANG Shu-zhi, HAYAT M, CAO Peng. Investigation of the deformation mechanism of a near β titanium alloy through isothermal compression [J]. *Metals*, 2017, 7: 498.
- [125] ZHANG Jing-qi, DI Hong-shuang, WANG Hong-tao, MAO Kun, MA Tian-jun, CAO Yu. Hot deformation behavior of Ti-15-3 titanium alloy: A study using processing maps, activation energy map, and Zener-Hollomon parameter map [J]. *Journal of Materials Science*, 2012, 47: 4000–4011.
- [126] YANG Qiu-yue, MA Min, TAN Yuan-biao, XIANG Song, ZHAO Fei, LIANG Yi-long. Initial β grain size effect on high-temperature flow behavior of TB8 titanium alloys in single β phase field [J]. *Metals*, 2019, 9: 891.
- [127] ZENG Li-ying, YANG Guan-jun, GE Peng, MAO Xiao-nan, ZHAO Yong-qing, ZHOU Lian. Processing map of one kind of metastable β titanium alloy [J]. *Rare Metal Materials and Engineering*, 2010, 39: 1505–1508.
- [128] ZHAO H Z, XIAO L, GE P, SUN J, XI Z P. Hot deformation behavior and processing maps of Ti-1300 alloy [J]. *Materials Science and Engineering A*, 2014, 604: 111–116.
- [129] BALASUBRAHMANYAM V V, PRASAD Y V R K. Deformation behaviour of beta titanium alloy Ti-10V-4.5Fe-1.5Al in hot upset forging [J]. *Materials Science and Engineering A*, 2002, 336: 150–158.
- [130] LYPCHANSKYI O, ŠLEBODA T, WOJTASZEK M, MUSZKA K, ŁUKASZEK-SOLEK A, STANIK R, GUDE M. The analysis of flow behavior of Ti-6Al-2Sn-4Zr-6Mo alloy based on the processing maps [J]. *International Journal of Material Forming*, 2021, 14: 523–532.
- [131] QU Wen-tao, SUN Xu-guang, HUI Song-xiao, WANG Zhen-guo, LI Yan. High-temperature deformation behavior of a beta Ti-3.0Al-3.5Cr-2.0Fe-0.1B alloy [J]. *Rare Metals*, 2018, 37: 217–224.
- [132] SARGENT P M, ASHBY M F. Deformation maps for titanium and zirconium [J]. *Scripta Metallurgica*, 1982, 16: 1415–1422.
- [133] PENG Xiao-na, GUO Hong-zhen, SHI Zhi-feng, QIN Chun, ZHAO Zhang-long, YAO Ze-kun. Study on the hot deformation behavior of TC4-DT alloy with equiaxed $\alpha+\beta$ starting structure based on processing map [J]. *Materials Science and Engineering A*, 2014, 605: 80–88.
- [134] LIANG Hou-quan, GUO Hong-zhen, NAN Yang, QIN Chun, PENG Xiao-na, ZHANG Jing-li. The construction of constitutive model and identification of dynamic softening mechanism of high-temperature deformation of Ti-5Al-5Mo-5V-1Cr-1Fe alloy [J]. *Materials Science and Engineering A*, 2014, 615: 42–50.
- [135] NING Y Q, XIE B C, LIANG H Q, LI H, YANG X M, GUO H Z. Dynamic softening behavior of TC18 titanium alloy during hot deformation [J]. *Materials & Design*, 2015, 71: 68–77.
- [136] NING Y Q, LUO X, LIANG H Q, GUO H Z, ZHANG J L, TAN K. Competition between dynamic recovery and recrystallization during hot deformation for TC18 titanium alloy [J]. *Materials Science and Engineering A*, 2015, 635: 77–85.
- [137] PRASAD Y V R K, RAO K, SASIDHARA S. Hot working guide: A compendium of processing maps [M]. 2nd ed. Materials Park, Ohio: ASM International, 2015.
- [138] ZHAO Qin-yang, YANG Fei, TORRENS R, BOLZONI L. Comparison of the cracking behavior of powder metallurgy and ingot metallurgy Ti-5Al-5Mo-5V-3Cr alloys during hot deformation [J]. *Materials (Basel)*, 2019, 12: E457.
- [139] ZHANG Jing-qi, DI Hong-shuang. Deformation heating and flow localization in Ti-15-3 metastable β titanium alloy subjected to high Z deformation [J]. *Materials Science and Engineering A*, 2016, 676: 506–509.
- [140] BALASUBRAHMANYAM V V, PRASAD Y V R K. Hot deformation mechanisms in metastable beta titanium alloy Ti-10V-2Fe-3Al [J]. *Materials Science and Technology*, 2001, 17: 1222–1228.
- [141] BOBBILI R, MADHU V. Constitutive modeling of dynamic flow behavior of Ti-5553 alloy [J]. *Journal of Alloys and Compounds*, 2019, 787: 260–266.
- [142] HUANG Bin, MIAO Xiao-fei, LUO Xian, YANG Yan-qing,

- ZHANG Yan-min. Microstructure and texture evolution near the adiabatic shear band (ASB) in TC17 Titanium alloy with starting equiaxed microstructure studied by EBSD [J]. Materials Characterization, 2019, 151: 151–165.
- [143] MORAKABATI M, HAJARI A. Hot working behavior of near alpha titanium alloy analyzed by mechanical testing and processing map [J]. Transactions of Nonferrous Metals Society of China, 2020, 30: 1560–1573.
- [144] LI C L, NARAYANA P L, REDDY N S, CHOI S W, YEOM J T, HONG J K, PARK C H. Modeling hot deformation behavior of low-cost Ti–2Al–9.2Mo–2Fe beta titanium alloy using a deep neural network [J]. Journal of Materials Science & Technology, 2019, 35: 907–916.
- [145] LONG Shuai, XIA Yu-feng, HU Jia-chen, ZHANG Jian-sheng, ZHOU Jie, ZHANG Peng, CUI Ming-liang. Hot deformation behavior and microstructure evolution of Ti–6Cr–5Mo–5V–4Al alloy during hot compression [J]. Vacuum, 2019, 160: 171–180.
- [146] LI X, LU S Q, FU M W, WANG K L, DONG X J. The optimal determination of forging process parameters for Ti–6.5Al–3.5Mo–1.5Zr–0.3Si alloy with thick lamellar microstructure in two phase field based on P-map [J]. Journal of Materials Processing Technology, 2010, 210: 370–377.
- [147] LEI Jia, ZHU Wen-guang, CHEN Lei, SUN Qiao-yan, XIAO Lin, SUN Jun. Deformation behaviour and microstructural evolution during the hot compression of Ti–5Al₄Zr₈Mo₇V alloy [J]. Materials Today Communications, 2020, 23: 100873.

高强 β 钛合金热变形过程中 组织演变及热加工图研究进展

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摘 要: 高强度 β 钛合金广泛应用于航空领域的大型承重部件。目前, 大型零件一般采用模锻成型。不同的初始显微组织和变形工艺参数会显著影响流动行为。为了实现显微组织的精确控制, 研究人员进行了大量的研究工作去分析热压缩过程中的组织演变和变形机制。本文重点综述高强度 β 钛合金在热变形过程中的组织演变, 包括单相区的动态再结晶和动态回复, 以及两相区中 α 相的动态演变。此外, 总结热加工图中的最佳热加工区域、不稳定区域以及功率耗散效率与变形机制之间的关系。最后, 强调利用热加工图优化工艺参数存在的问题和发展方向。

关键词: 高强 β 钛合金; 热变形; 组织演变; 热加工图

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