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Hot deformation behaviours and spheroidization mechanisms of Ti-5322 alloy during hot compression

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Abstract

The hot deformation behavior of Ti-5322 alloy are researched at compression temperatures range of 750–1050°C and strain rate range of 0.01–10 s⁻¹, to optimize its hot workability. Processing map analysis and microstructure observations reveal that the optimal processing parameters of Ti-5322 alloy are temperatures of 750–825°C and strain rates of 0.01–0.05 s⁻¹, and temperatures of 925–975°C and strain rates of 0.01–1 s⁻¹. The peak efficiency of power dissipation can reach 40% owing to the transformation from α phase to β phase, spheroidization behavior and dynamic recrystallization of the β phase. The dynamic recrystallization was the primary form of microstructure evolution above 900°C, while the spheroidization of α phase below 900°C. The spheroidization of α lamellae can be attributed to the instability of subgrain boundaries appeared in the α phase during hot deformation. The β phase wedges into the α/α subgrain boundary and α/β interface migration induced the α phase spheroidization. In addition, three instability domains are detected in the processing maps, which confirmed by the presence of microstructures with wedge cracking and adiabatic shear bands.

1. Introduction

Ti-5322 alloy is a new Ti-Fe-V-Cr-Al alloy with high hardenability, high strength, and excellent application potential. It is designed to be inexpensive and to have good recyclability, excellent hot workability, and outstanding heat treatment strengthening. In addition to reducing the raw material cost, the formulation for hot working also affects the cost of new materials. Furthermore, the application of computer simulation technology in materials research has become more prevalent. The microstructure evolution can be predicted by computer simulation, which can reduce the number of repeated experiments. Therefore, computer simulation has become an important technology for controlling costs in the research and development of new materials [1]. Philippart and Rack [2] investigated the micro-structure evolution of Ti-6.8Mo-4.5Fe-1.5Al near-β alloy at hot compression. The stability domains were found to correspond to dynamic recovery (DRV) at low temperatures and dynamic recrystallization (DRX) at high temperatures. Furthermore, wedge cracking and adiabatic shear bands can be noticed in the instability domains. Yu Sun et al [3] investigated the deformation mechanism of Ti-22Al-25Nb alloy at different deformation parameters, and found that the optimal deformation region of the alloy occurred at low strain rates and high temperature. Wu et al [4] established the DRX model and constitutive equation of the Ti55531 alloy during hot deformation. Maet al [5] reported spheroidization of α lamellae in BT25 titanium alloy and discovered that the volume fraction of spheroidized α lamellae raised with deformation degree. Meanwhile, the lamellar termination migration behavior became more active with increasing temperature, which leading to dynamic spheroidization to occur more easily. Most research on Ti-5322 alloy to date has been limited to studying the effects of conventional rolling deformation on the properties. Few papers on the spheroidization mechanisms and processing maps of Ti-5322 alloy have been reported. Therefore, the systematic investigation of the microstructure transformation during hot deformation is vital for optimizing the hot workability and controlling the microstructure evolution, as such research can provide guidance for process control and defect prediction during hot deformation.
The primary purposes of this study are to construct constitutive models of Ti-5322 alloy, and analyze the effects of different compression experiments parameters for the flow stress, as well as the processing maps are established. In this study, the stability and instability domains during deformation based on analyzing both the processing maps and the microstructure evolution are confirmed. Furthermore, the spheroidization mechanisms of \( \alpha \) lamellae during hot working are investigated using transmission electron microscopy (TEM).

2. Material and experiments

In this experiment, the Ti-5322 alloy was used, and its chemical composition (wt\%) of 4.69Al-1.91Cr-2.62V-1.78Fe-0.15O-Ti (bal). The \( \beta \) transus temperature of the Ti-5322 alloy is about 905 \( ^\circ\)C, as measured by differential thermal analysis. In contrast to the conventional TC4 alloy, 1.78 wt\% Fe replaces some of the V and other precious metal elements. Additionally, the Al content is slightly lower, and 0.15 wt\% O is added to harden the alloy.

The initial samples were cylindrical bars (8 mm \( \times \) 12 mm) produce by a linear cutting. Figure 1(a) displays the initial microstructures of Ti-5322 alloy. It can be noticed that microstructures are composed of equiaxed \( \alpha \) phase and \( \beta \) phase, and the \( \alpha \) phase is uniformly distributed on the \( \beta \) transformation matrix. Meanwhile, in order to obtain lamellar \( \alpha \) phase, the initial samples were annealed at 920 \( ^\circ\)C for 20 min, followed by air cooling. After heat treatment, it can be found a number lamellar \( \alpha \) phase appear in \( \beta \) matrix, as presented in figure 1(b). The compressive deformation experiment was performed on Gleeble-3500 testing machine. In order to eliminate the friction between the specimen and the experiment mold, the tantalum sheets were set on both ends of the sample. Also in order to reduce the negative effects of uneven temperature distribution, the temperature was heated to the deformation temperature at a slow heating rate of 10 \( ^\circ\)C/s, and insulate for 5 min. The experiment in this paper is mainly divided into two parts. The first part is mainly to study the deformation behavior of Ti-5322 alloy with an equiaxed microstructure. The selected deformation parameters include seven deformation temperatures (750 \( ^\circ\)C, 800 \( ^\circ\)C, 850 \( ^\circ\)C, 900 \( ^\circ\)C, 950 \( ^\circ\)C, 1000 \( ^\circ\)C and 1050 \( ^\circ\)C), four strain rates (0.01 s\(^{-1}\), 0.1 s\(^{-1}\), 1 s\(^{-1}\) and 10 s\(^{-1}\)) and height reduction 40%. The second part focused on the spheroidization mechanisms of \( \alpha \) lamellae during hot deformation; the initial microstructure of the samples should be in the form of lamellar. Therefore, the initial microstructure of the specimens is lamellar \( \alpha \) phase, as show the figure 1(b). The deformation parameters is 850 \( ^\circ\)C, two strain rates (0.01 s\(^{-1}\), 0.1 s\(^{-1}\)), and height reduction 40%.

After the hot deformation tests, all the specimens were quenched immediately to room temperature. The axial sections of the central portions of the deformed specimens. Then, the deformation specimens were manual polishing and chemically etched with a solution of 10\% HF+20% HNO\(^3\)+70%H\(_2\)O. Finally, the microstructure was investigated by optical microscopy (ZEISS Axiovert) and TEM (Tecnai F30 G2).

3. Experimental results and discussion

3.1. Flow behaviours and deformation characteristics

Figure 2 displays the change of the deformation flow stress in isothermal compression with different deformation parameters, and the stress-strain curves are considered to be obtained with less friction, so the given stress-strain curves were not corrected. The flow stress is extremely affected by processing parameters. The true stress declines with the increasing deformation temperature. The reason is that the effective stress required for a dislocation and the atomic binding force is reduced at higher deformation temperatures [6]. Macroscopic-ally, this behavior indicates...
that the critical value for crystal slip decrease and the number of movable slip systems increase, which reduces the resistance of dislocation motion and dislocation slip, and induced the decrease of the flow stress. Additionally, the hot deformation process include DRX and DRV, the DRX and DRV take place with raised temperature. It causes the dislocation density in the alloy to decrease, possibly eliminating the work hardening and producing a strain softening effect.

Figure 2 depict the flow stress are significantly affected by strain rate. when the strain rate is changed from 0.001 s$^{-1}$ to 10 s$^{-1}$, the peak stress obviously raise. This is because the DRX and DRV associated with diffusion cannot be fully completed at high strain rates, the dislocation motion becomes difficult and the dislocation density sharply increased, which promotes the degree of work hardening and peak stress. Contrarily, dynamic recovery and dynamic recrystallization can be fully completed at low strain rates, which leading to decreases of the dislocation density and flow stress.

3.2. Constitutive equations

The plastic deformation of a metal is a complicated process. According to previous research, the relationships between the flow stress and deformation parameters can be constructed using the Arrhenius constitutive model with an apparent activation energy Q$^{[7,8]}$. They are expressed as

\[
\dot{\varepsilon} = A_1 \sigma^{\alpha} \exp \left( -\frac{Q}{RT} \right) (\alpha \sigma < 0.8) \tag{1}
\]

\[
\dot{\varepsilon} = A_2 \exp (\beta \sigma) \exp \left( -\frac{Q}{RT} \right) (\alpha \sigma > 1.2) \tag{2}
\]

\[
\dot{\varepsilon} = A [\sin h(\alpha \beta)]^{n} \exp \left( -\frac{Q}{RT} \right) (\forall \sigma \sigma) \tag{3}
\]

where A, α, and n1 are material constants; \(\alpha = \beta/n1\), R = 8.314 J(molk)$^{-1}$, \(\dot{\varepsilon}\) is the strain rate, and \(\sigma\) is the flow stress.

Figure 2. Flow stress curves of Ti-5322 alloy at (temperatures: 750 °C, 800 °C, 850 °C, 900 °C, 950 °C, 1000 °C and 1050 °C): (a) 0.01 s$^{-1}$, (b) 0.1 s$^{-1}$, (c) 1 s$^{-1}$, and (d) 10 s$^{-1}$.
The effects of experiment parameters for the hot compression process of materials can be represented by an exponential equation of the Zener–Hollomon (Z) parameter [9]:

\[
Z = \dot{\varepsilon} \exp\left(\frac{Q}{RT}\right)
\]

Thus, the relationship between \(\sigma\) and \(Z\) can be obtained by combining equations (3) and (4), as shown in equation (5):

\[
\sigma = \frac{1}{\alpha} \ln \left\{ \left[\frac{(Z/A)^{1/n} + [(Z/A)^{2/n} + 1]^{1/2}}{\sinh(\alpha\sigma)}\right] \right\}
\]

The material constants such as A, \(\alpha\), n, and Q can be obtained by analyzing the true stress–strain curves acquired by deformation.

To reflect the compression process of Ti-5322 alloy more realistically, the constitutive equations of Ti-5322 alloy are established. The flow stress corresponding to a true strain of 0.2 is used to illustrate the method of deriving the parameters required for the Arrhenius constitutive model. This is because after the stress reaches the peak value, which reduces with the raised of the strain. From figure 2, When the strain increases to 0.2, indicating the effects of work hardening are weakened by softening mechanisms, the stress tends to be balanced state. The values of \(n_1\) and \(\beta\) can be determined by the \(\ln \sigma - \ln \dot{\varepsilon}\) and \(\ln \dot{\varepsilon},\) as shown in figures 3(a) and (b), respectively. The \(\ln \sigma - \ln \dot{\varepsilon}\) and \(\ln \dot{\varepsilon}\) curves both show good linear correlations at different temperatures, whereas the slopes of these lines vary significantly with the region. The values of \(n_1\) and \(\beta\) in different phase regions are calculated as 5.62149 and 0.04331, and 3.97362 and 0.06123, respectively. Using the average slopes of the straight lines in figures 3(a) and 3(b).

Assuming that the apparent activation energy (Q) can be independent of T in a certain temperature range, the former is defined as

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

The value of Q can be confirmed by calculating the slopes of the plots shown in figures 3(c) and (d), respectively. The apparent activation energy at a strain of 0.2 in the double-phase and single-phase regions of Ti-5322 alloy is found to be 397.66 and 244.24 KJ mol\(^{-1}\), respectively.
After the apparent activation energy is determined, Z can be calculated using the following formula:

\[
\ln Z = \ln A + n \ln (\sinh (\alpha \sigma))
\]

From figure 4 Plots of \( \ln Z \) versus \( \ln (\sinh (\alpha \sigma)) \) for different regions. From figure 4(a), \( n \) is calculated to be 3.69, and \( \ln A \) is 40.64 in the double-phase region. Figure 4(b) shows the value of \( n \) is 2.93, and the \( \ln A \) is 23.49 in the single-phase region.

Then, substituting the relevant parameters into equation (5), the flow stress in the different phase region can be predicted using the expressions containing Z in equations (8) and (9):

\[
\sigma = \frac{1}{0.0077} \ln \left\{ \left( \frac{Z}{4.45 \times e^{1.7}} \right)^{1/3.69} + \left( \frac{Z}{4.45 \times e^{1.7}} \right)^{2/3.69} + 1 \right\}^{1/2}
\]

\[
\sigma = \frac{1}{0.0154} \ln \left\{ \left( \frac{Z}{1.59 \times 10^{10}} \right)^{1/2.93} + \left( \frac{Z}{1.59 \times 10^{10}} \right)^{2/2.93} + 1 \right\}^{1/2}
\]

3.3. Processing maps

To apply the constitutive model of materials using finite element analysis, Prasad and Seshacharyulu [10] proposed a dynamic material model based on the fundamental principles of large plastic deformation continuum mechanics, physical system models, and irreversible thermodynamics. This approach can be seen as a bridge connecting large plastic deformation continuum mechanics with the material dissipative structure. The flow stress of a strain-rate-sensitive material is generally expressed in the form of Backoken’s relation:

\[
\sigma = K \cdot \dot{\varepsilon}^m
\]

where \( m \) is the strain rate sensitivity coefficient, \( K \) is constant.

The total energy input into the artifact can be expressed as

\[
P = \sigma \dot{\varepsilon} = G + J = \int_0^t \sigma d\dot{\varepsilon} + \int_0^\sigma \varepsilon d\sigma
\]

where \( G \) is the dissipator content caused by plastic deformation, most of which is converted into heat, and \( J \) is the dissipator co-content, which dissipates as the microstructure evolution [10]. The strain rate sensitivity coefficient \( m \) affects the power distribution between \( G \) and \( J \), as given below [10]:

\[
m = \frac{\partial J}{\partial G} = \frac{\varepsilon \partial \sigma}{\sigma \partial \varepsilon} = \frac{\partial (\ln \sigma)}{\partial (\ln \dot{\varepsilon})}
\]

Thus, the dissipator co-content \( J \) is defined as

\[
J = \int_0^\sigma \varepsilon d\sigma = \frac{m \sigma \dot{\varepsilon}}{1 + m}
\]
The value of $\eta$ can be obtained as

$$\eta = \frac{f}{f_{\text{max}}} = \frac{2m}{1 + m} \quad (14)$$

where $f_{\text{max}} = \sigma \dot{\varepsilon}/2 = P/2$. When $m = 1$, the material at an ideal linear dissipation state.

In 1965, Ziegler [11] proposed an instability condition for plastic deformation:

$$\frac{dD}{d\dot{\varepsilon}} < \frac{D}{\dot{\varepsilon}} \quad (15)$$

where $D(\dot{\varepsilon})$ is a function of the dissipation. When the total energy is divided into the dissipation $G$ and dissipation covariant $J$, the function $D(\dot{\varepsilon})$ can be replaced by the co-content $J$, as shown below:

$$\frac{dj}{d\dot{\varepsilon}} < \frac{J}{\dot{\varepsilon}} \quad (16)$$

In 1998, Prasad and Seshacharyulu derived a flow instability criterion associated with the microstructure, which was defined as:

$$\xi(\dot{\varepsilon}) = \frac{\partial \ln \left( \frac{m}{m + 1} \right)}{\partial \ln \dot{\varepsilon}} + m < 0 \quad (17)$$

The $\eta$ and the instability coefficient $\xi$ are both functions of the $m$, suggesting that the value of $m$ is one of the crucial factors for establishing a processing map. To improve the accuracy of the calculation, cubic spline function fitting is applied to $\ln \sigma$ and $\ln \dot{\varepsilon}$, as follows:

$$\ln \sigma = a + b(\ln \dot{\varepsilon}) + c(\ln \dot{\varepsilon})^2 + d(\ln \dot{\varepsilon})^3 \quad (18)$$

The value of $m$ can be calculated according to equation (12):

$$m = b + 2c(\ln \dot{\varepsilon}) + 3d(\ln \dot{\varepsilon})^2 \quad (19)$$

The change of $\eta$ with deformation parameters constitutes the power dissipation maps according to equation (14).

Similarly, the relationship between $\ln (m/1 + m)$ and $\ln \dot{\varepsilon}$ can be expressed as

$$\ln (m/1 + m) = k_1 + k_2(\ln \dot{\varepsilon}) + k_3(\ln \dot{\varepsilon})^2 + k_4(\ln \dot{\varepsilon})^3 \quad (20)$$

The instability parameters $\xi(\dot{\varepsilon})$ can be obtained according to equation (17):

$$\xi(\dot{\varepsilon}) = k_2 + 2k_3(\ln \dot{\varepsilon}) + 3k_4(\ln \dot{\varepsilon})^2 + m < 0 \quad (21)$$

The change of the instability parameter $\xi(\dot{\varepsilon})$ with deformation parameters constitutes the instability map. The processing maps at true strains of 0.2 and 0.4 are obtained by superimposing the power dissipation maps and the instability map, as shown in figures 5(a) and (b), respectively. The contour lines and shaded regions represent the power dissipation efficiency and instability domains, as shown in figure 5, respectively.

3.3.1. Stability domains

As shown in figure 5, the value of $\eta$ decreases with raised strain rate and reduced experiment temperature. Generally, the value of $\eta$ indicates the power dissipation capacity of the material and microstructure transformation mechanism. The higher $\eta$ value represents that a higher proportion of the power was applied to microstructure evolution during isothermally compressed. Meanwhile, the hot workability of a materials affects by the value of $\eta$. Therefore, the microstructures corresponding to the maximum and minimum $\eta$ values in the processing map were analyzed, and to research the effect of different experiment parameters on the microstructure evolution. However, the areas with higher $\eta$ values are not always safe regions; thus, it is necessary to combine the flow instability criterion to identify the instability domain.

As shown in figures 5(a) and (b), the suggested deformation region of the processing maps at different true strains is represented by dotted lines. In addition, figures 5(a) and (b) demonstrate the optimal deformation regions of Ti-5322 alloy are 750°C–825°C, 0.01–0.05 s⁻¹; 925°C–975°C, 0.01–1 s⁻¹; and 1000°C–1050°C, 0.05–1 s⁻¹. In these regions, hot deformation is more efficient, and the value of $\eta$ is 0.35–0.49. According to the reported [12], the materials occurrence of DRX, DRV and superplastic deformation in the stability domains. As shown in figures 6(a)–(d), the deformed microstructure at different deformation parameters. From figures 6(b)–(d), the α lamellae become shorter or thicker, and no dynamic recrystallized grains can be observed. Meanwhile, many spheroidized structures appear in the β matrix. In addition, the microstructure become fine and uniform, indicating that lamellate α phase are spheroidized at these conditions. Therefore, analysis of the spheroidized mechanism of lamellar α phases is significant for optimizing the process parameters and will be discussed in
detail below. Figures 6(e)–(h) shows that a number of dynamically recrystallized grains appear in the β phase boundaries and interiors.

From figures 6(e), (g), when the strain stress is 0.1 s⁻¹, the increased deformation temperature can promote the growth of β grain. The reason is increased temperature benefit the elements diffusion, and declines the fraction of α phases [12].

Figures 6(f), (g) demonstrates different strain rates significantly affects microstructure of β grain. When the strain rate is raised from 0.01 s⁻¹ to 0.1 s⁻¹, it is observed that the DRX of grain is increased at 950 °C, because the higher strain rate can not provide abundant time for dislocation motion, which result in the dislocation rapidly multiplication and induce the transformation from low angle boundaries into the high angle grain boundaries [6]. Therefore, the deformation mechanism of Ti-5322 alloy is affect by the DRX of the β phase.

3.3.2. Instability domains

The variation in ξ with deformation parameterse provides an instability map in which the instability domains are characterized by negative ξ values. Figure 7 displays the instability parameter for the hot deformation Ti-5322 alloy with microstructure evolution at different compression parameters. In figures 7(a) and (b), the flow instability is restricted to three main regions. Also the value of η is generally less than 30%. This result indicates that less power is applied to improve the microstructure in these instability domains, and most of it is applied to plastic deformation of the alloy and eventually converted into heat. Thus, these domains should be avoided during deformation of Ti-5322 alloy.

The stability during hot deformation of metals mainly relevant to wedge cracking, localized plastic flow, and adiabatic shear bands. Therefore, observing the microstructure features of the instability domains is essential for
illustrating the primary instability mechanism of Ti-5322 alloy during hot deformation. As shown in figure 7(a), wedge cracking occurs in the hot deformation of Ti-5322 alloy at this time. The generation of wedge cracking is mainly relevant to the internal stress, and the internal stress is caused not only by external forces, but also by inhomogeneous deformation, an uneven temperature distribution, and asynchrony in the structural transformation in plastic ability during hot deformation of metals is probably relevant to deformation defects. Therefore, observing the microstructure of the instability domains is essential for illustrating the primary instability mechanism of Ti-5322 alloy during hot deformation. As shown in figure 7(a), the wedge cracking occurs in the hot deformation of Ti-5322 alloy, the microscopic cracks appear near the shear bands. The generation of wedge cracking is related to the internal stress, and the internal stress is caused not only by external
forces, but also by inhomogeneous deformation, an uneven temperature distribution, and asynchrony in the structural transformation in plastic deformation. Cracking is generated in a material when the internal stress exceeds a threshold.\[13\] Figure 7(b) and figure 5, it is found significant adiabatic shear band in the deformed sample, and the value of \(\eta\) is generally less than 30%, most of which during the alloy deformation process is used for plastic deformation. Compared to conventional metals, Ti-5322 alloy has the disadvantage of a smaller thermal conductivity coefficient. The thermal energy produced by plastic deformation cannot be sufficiently released by heat diffusion under high strain rate conditions. Therefore, the temperature increases in local areas, resulting in an intense flow softening effect, and the adiabatic shear band occurs more easily in these areas. In the shear deformation zone, the shear deformation is very concentrated. In many cases, the adiabatic shear band is often a region where cracks initiate and expand, causing flow instability of the material during hot deformation.

3.4. Mechanisms of spheroidized

Ti-5322 titanium alloy has two allotropes. At high temperature region above the phase transition point, the alloy consists of a single \(\beta\) phase. Blow the phase transition point, the alloy consists of \(\alpha\) and \(\beta\) phase. The process of titanium alloys is generally generate an equiaxed structure with great tensile and fatigue properties. However, the lamellar structure obtained by rapid cooling is stable, and it can only be equiaxed by radical deformation in two-phase region. During the hot compression, the orientation of the spheroidized \(\alpha\) phase variations randomly, and the spheroidized \(\alpha\) phase is preferentially formed near the equiaxed \(\alpha\) phase.\[14\]

The transformation process of the lamellar structure to equiaxed structure is complicated when deformed in two-phase region. Li L.\[15\] suggested that dynamic recrystallization is a major factor affecting the spheroidization of \(\alpha\) lamellae. Seshacharyulu et al.\[16\] and Stefansson and Semiatin\[17\] found that lamellar structure is spheroidized by shear bands passing through the \(\alpha\) lamellae. However, Weiss et al.\[18\] attributed the spheroidization of \(\alpha\) lamellae to the substructure produced by dynamic recovery. In order to research and analyze the microstructure evolution of lamellar structure of Ti-5322 titanium alloy during hot forming, the deformed structure was analyzed by TEM. It was found that the spheroidization mechanism of \(\alpha\) lamellar structure mainly includes the following two types.

3.4.1. Shear deformation mechanism

The separation of \(\alpha\) lamellae is revealed by TEM observations, as presented figure 8. Figure 8(a) illustrates the slip lines exist in the \(\alpha\) lamellae and penetrate entire \(\alpha\) lamellae, generating a slip step. The no obvious slip lines can be observed at certain deformation conditions, but the slip step is significant on the wide side of the \(\alpha\) lamellae, as shown in figure 8(b). From figure 8(c), it is found the number of dislocation are occurrence near the slip bands, and some \(\alpha\) phase became fragmentation. It is attributed to the plastic deformation promote the broken of \(\alpha/\beta\) interface and separation of \(\alpha\) lamellae. Form figure 8(d), it can be noticed the thermal etching groove is continuously deepened with the subgrain boundaries are formed in \(\alpha\) phase, until the \(\alpha\) lamellae is completely split. The separation of \(\alpha\) lamellae occurs by atomic diffusion of the \(\beta\) phase into the \(\alpha\) phase, as presented in figure 8.

Figure 7. Microstructures of instability regions: (a) wedge cracking microstructure at 850 °C, 10 s\(^{-1}\), (b) adiabatic shear band microstructure at 1000 °C, 10 s\(^{-1}\).
3.4.2. Dynamic recovery substructure mechanism
The α phase of titanium alloy has a higher stacking fault energy, thus, dynamic recovery occurs more easily during hot deformation. The dislocation multiplication caused by plastic deformation increases the number of movable dislocations. Dislocation walls and subgrain boundaries are formed by dislocation slip, cross-slip, and climbing of movable dislocations, as shown in Figure 9(a). From Figure 9(b), the multiple subgrain boundaries formed in the same α lamella phase induces the stability of the α lamella phase decrease, the α lamella phase is divided into several segments, and the thermal etching groove is formed at the intersection of the α/β interface and α/α subgrain boundary as shown in Figure 9(c). The depth and width of the thermal etching groove gradually increase with increase of atomic diffusion into the α/α subgrain boundary. Meanwhile, the β phase wedges into α lamella phase along the subgrain boundary, the α lamella phase is fragmented [19]. Finally, the α lamellae is split into some spheroidized α phase, as shown in Figure 9(d).

The spheroidization process of lamellar structure is shown in Figure 10. The tension between an α/α subgrain boundary and an α/β interface is in the non-equilibrium state at the intersection of the α/β interface and the subgrain boundary formed by plastic deformation, as shown in Figure 10(a). As shown in Figure 10(b), a thermal etching groove is formed by diffusion of the β phase into the α/α subgrain boundary to balance the interfacial tension. From Figure 10, the relationship between interface energy γα/α and phase interface energy γα/β can be obtained as:

\[ \gamma_{\alpha/\alpha} = 2 \gamma_{\alpha/\beta} \cos (\theta / 2) \]  \hspace{1cm} (22)

where \( \theta \) is the angle of thermal etching groove. From the formula (22), it can be found that the higher the subgrain boundary energy γα/α or the larger the grain boundary angle, the smaller the \( \theta \) and the deeper the β phase wedged into α lamellae. As shown in Figure 10, assuming that the depth of the thermal etching groove is \( H \), and the distance between the tangent of the α/β interface and the intersection of the lamellae surface is \( 2L \), it can be obtained from equation (22):

\[ H^2 = \frac{\gamma_{\alpha/\alpha}^2}{4 \gamma_{\alpha/\beta}^2 - \gamma_{\alpha/\alpha}^2} L^2 \]  \hspace{1cm} (23)

Generally, the γα/β is a constant. Therefore, the depth of β phase wedging into α phase mainly depends on interfacial energy γα/α and the width of thermal etching groove 2 L. The H value increases with the increasing interface energy γα/α. When the interface α/α is a low-angle grain boundary (\(<15^\circ\) ), the interface energy γα/α increases with increasing interface angle. However, when it translates into a high-angle grain boundary (\(>15^\circ\) ),...
the interface energy $\gamma_{\alpha/\alpha}$ is almost independent of the grain boundary angle. This displays that the increase of the grain boundary angle may limited effect on the depth of the thermal etching groove. Therefore, the increase of the width of thermal etching groove is essential for the separation of $\alpha$ lamellae. The separation mechanism for $\alpha$ lamellae is attributed to the $\beta$ phase wedges into the $\alpha/\alpha$ subgrain boundary and $\alpha/\beta$ interface migration. Because there are curvature differences between lamellae at different locations, and atoms diffuse along the $\alpha/\beta$ interface under the Gibbs–Thomson effect [20], resulting in dissolution of the $\alpha$ phase near the bottom of the
thermal etching groove and an increase in the depth of diffusion of the $\beta$ phase into the $\alpha/\alpha$ subgrain boundary. Finally, the $\beta$ phase penetrates the $\alpha$ lamellae, causing separation and spheroidization of the $\alpha$ lamellae [16, 19].

The spheroidization of $\alpha$ lamellae is a combined effect of thermal and mechanical processes. Various mechanisms have been proposed to explain this phenomenon. On the basis of previous research, Zherebtsov et al demonstrates that the lamellar $\alpha$ phase of titanium alloy is spheroidized by two mechanisms during hot deformation [21]. One mechanism is the fragmentation and spheroidization of $\alpha$ lamellae caused by interface separation, which requires the generation of subgrain boundaries in the $\alpha$ lamellae, the subgrain boundaries can be formed by dynamic recrystallization, dynamic recovery, and localized shear bands. The other mechanism is termination migration, which results mainly in diffusion coarsening and spheroidization of $\alpha$ lamellae during heat treatment. The driving forces of termination migration are the chemical potential and curvature differences between lamellae at different locations [22]. The theory is consistent with the experimental results, which further validates the analyses of the spheroidization mechanism of lamellar structure in this paper.

4. Conclusions

In summary, the flow stress during isothermal compression of Ti-5322 alloy is greatly affected by the different deformation parameters. In brief, the flow stress drops with raise deformation temperature or the declines strain rate. According to the Arrhenius-type constitutive equation, the constitutive models of the single-phase and double-phase regions of Ti-5322 alloy are constructed, the expression is as follows:

$$\sigma = \frac{1}{0.0077} \ln \left\{ \left( \frac{Z}{4.45 \times e^{17}} \right)^{1/3.69} + \left[ \frac{Z}{4.45 \times e^{17}} \right]^{1/3.69} + 1 \right\}^{1/2}$$

$$\sigma = \frac{1}{0.0154} \ln \left\{ \left( \frac{Z}{1.59 \times 10^{10}} \right)^{1/2.93} + \left[ \frac{Z}{1.59 \times 10^{10}} \right]^{1/2.93} + 1 \right\}^{1/2}$$

It is found the optimal processing parameters of Ti-5322 alloy are $750–825 \degree C, 0.01–0.05 s^{-1}; 925–975 \degree C, 0.01–1 s^{-1};$ and $1000–1050 \degree C, 0.05–1 s^{-1}$. Microstructure observations reveal that most $\alpha$ lamellae are spheroidized in the double-phase region, and dynamic recrystallization is the principal type of microstructure evolution above $900 \degree C$. Meanwhile, the spheroidization of $\alpha$ lamellae can be attributed to the instability of subgrain boundaries formed in the $\alpha$ phase during hot deformation. Internal defects are formed via subgrain boundaries caused by shear deformation and substructure generated by DRV.

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