Effect of strain rate and deformation temperature on strain hardening and softening behavior of pure copper

Shu-hai HUANG, Da-yu SHU, Chuan-kai HU, Shi-feng ZHU
Southwest Technique and Engineering Institute, Chongqing 400039, China
Received 21 April 2015; accepted 3 December 2015

Abstract: The effects of the deformation temperature and the strain rate on the hot deformation behavior of pure copper were investigated based on compression tests. The expressions of strain hardening rate, dynamic recrystallization critical stress, saturated stress, dynamic recovery volume fraction and dynamic recrystallization volume fraction were determined. According to the processing map, the instability regions occur in regions of $400-450^\circ C$, $0.001-0.05$ s$^{-1}$ and $450-750^\circ C$, $0.05-1$ s$^{-1}$. The deformation mechanism in the stability region is dynamic recrystallization. The flow stress was predicted. The results also show that the true stress–true strain curves predicted by the extracted model are in good agreement with the experimental results.

Key words: copper; hot compression deformation; strain hardening; strain softening

1 Introduction

Pure copper cartridge receivers with complex geometries used in the ordnance industry are generally manufactured at high temperatures and very low strain rates due to the strict requirement of the fine-grained structure. Forging at high temperatures is widely used to obtain copper parts with complex shapes. Therefore, to optimize the fabrication process, it is necessary to study the hot deformation behavior of pure copper at high temperatures and low strain rates. Hot deformation behavior is often simulated by compression tests due to their similarity to forging processes. When pure copper is compressed at high temperatures, the flow stress curves present three typical regions of hot deformation: a strain hardening (SH) region characterized by an increasing dislocation density, followed by a dynamic recovery (DRV) characterized by polygonal sub-grain structures, and finally dynamic recrystallization (DRX) in which build-up of dislocations leads to nucleation and growth of DRX grains during the deformation process until a steady state is reached [1–3].

The hot deformation behaviors of nonferrous metals, such as constitutive equations, prediction of flow stress, flow softening behaviors, mainly focused on aluminum alloys, magnesium alloys and titanium alloys [4–7]; however, investigations on strain hardening and softening behaviors of pure copper are very rare. GRONOSTAJSKI [8] proposed a general model to describe the flow stress as a function of the internal state of copper alloys. The model described strain hardening and dynamic softening processes taking place during the deformation process or static softening during interval or after deformation. MANONUKUL and DUNNE [9] established a hot deformation model of pure copper. The model was calibrated using the homogeneous test data and used to predict the deformation, the initiation of DRX, and the microstructural change in the truncated cone specimens. After comparing the computed results with the experimental results, the model was proven to be capable of tracking the motion of recrystallization fronts through the deforming material under conditions of inhomogeneous stresses. PRASAD and RAO [10] studied mechanisms of high temperature deformation in electrolytic copper in extended ranges of temperature and strain rate. The switching of the rate controlling mechanisms at $700^\circ C$ was attributed to the “clogging” of the dislocation pipes due to the rapid increase in the solid solubility of oxygen at high temperatures.

The aim of this work is to study the effects of strain rate and deformation temperature on strain hardening and softening behaviors of pure copper. Therefore, we can gain a better understanding of hot forging for pure
copper cartridge receivers with complex geometries.

2 Experimental

The material used in this work was pure copper bar, which was annealed at 420 °C for 2 h. Cylindrical specimens with 12 mm in height and 8 mm in diameter were cut from the annealed bar, and the initial average size was about 180 μm. Compression test was carried out on a Gleeble−1500 thermal simulator in the temperature range of 400−900 °C and strain rate range of 0.001−1 s⁻¹. The total compression deformation was 60%. The whole compression process was controlled by a computer capable of collecting relevant data in an automatic way.

Figure 1 shows the stress−strain curves of pure copper under different conditions. Typically, the flow stress decreases with increasing the temperature and decreasing the strain rate. When the deformation temperature is comparatively low, strain hardening dominates the deformation, which is more apparent in the high strain rate region. After peak strain, the flow stress decreases with the increase of strain; with further

Fig. 1 Stress−strain curves of pure copper at different strain rates and temperatures: (a) 400 °C; (b) 500 °C; (c) 600 °C; (d) 700 °C; (e) 800 °C; (f) 900 °C
increase in the strain, the secondary strain hardening takes place. Compared with the first strain hardening rate, the secondary hardening rate is much lower. The shape of the stress–strain curves can reflect the mechanisms of hot deformation. For example, the flow stress increases with increasing the strain in the initial stage of deformation, which is always caused by the strain hardening. Then, the flow stress decreases due to the dynamic recrystallization or recovery processes. Detailed analysis about the hot deformation mechanism will be carried out in the subsequent discussion.

3 Principles of strain hardening and softening

During thermoplastic deformation, the variation of flow stress is generally determined by strain hardening and softening [11–14]. Plastic deformation is primarily completed by dislocation movement, multiplication and annihilation. The dislocation density increases with the increase of strain, resulting in the strain hardening. There are two kinds of mechanisms of strain softening: dynamic recovery and dynamic recrystallization. The behavior of flow stress can be roughly divided into three stages (Fig. 2). Firstly, before strain \( \varepsilon \) reaches the initial yield strain, \( \varepsilon_0 \), i.e., \( \varepsilon < \varepsilon_0 \), the elastic deformation of materials is dominant. Secondly, before the strain \( \varepsilon \) approaches the critical strain \( \varepsilon_c \) of dynamic recrystallization, i.e., \( \varepsilon_0 < \varepsilon < \varepsilon_c \), plastic deformation of materials takes place. In this situation, strain hardening and dynamic recovery are dominant. At last, when the strain \( \varepsilon \) reaches an steady state deformation \( \varepsilon_{ss} \), i.e., \( \varepsilon_c < \varepsilon < \varepsilon_{ss} \), strain hardening, dynamic recovery and dynamic recrystallization softening occur alternately with the increase of the deformation degree. Peak strain \( \varepsilon_p \) can be considered as the initial point of the balance that is achieved between the strain hardening posed by dislocation pile-up and the recovery softening resulted from the dynamic recrystallization. Assuming that there is no dynamic recrystallization during deformation and only the strain hardening and the dynamic recovery are available, the dynamic recovery stress (\( \sigma_{rec} \)) curve may be obtained by calculation. \( \sigma_{sat} \) is the asymptotic stress under the state of saturated stress.

3.1 Modeling strain hardening and dynamic recovery

When the deformed material turns into the state of yield, the plastic deformation of materials becomes obvious. A large number of dislocation movements take place inside the materials. Consequently, interactions including initiation of fixed dislocation movement, dislocation multiplication, annihilation and rearrangement are enhanced. Dislocation density, \( \rho \), is primarily determined by the increased dislocation density induced by the strain hardening during the deformation, while the reduced dislocation density is resulted from the dynamic recovery [11–13]. The dynamic recovery stress \( \sigma_{rec} \) of the material can be written as

\[
\sigma_{rec} = \left[\sigma_{sat}^2 - (\sigma_{sat}^2 - \sigma_0^2) \exp[-r(\varepsilon - \varepsilon_0)]\right]^{1/2}
\]

where \( \sigma_0 \) is the yield stress, \( \varepsilon_0 \) is the initial yield deformation, and \( r \) is the parameter of DRV.

3.2 Modeling dynamic recrystallization critical deformation

In the modern theory of dynamic recrystallization, the relation of \( \varepsilon_c = \alpha_D \varepsilon_p \) is generally used to determine the critical strain \( \varepsilon_c \) of dynamic recrystallization, where \( \alpha_D \) is the constant parameter. In the process of hot deformation of pure copper, various interactions of dislocation mechanisms, hardening, softening, diffusion activation and effective stress constitute a compound movement of dislocation during the deformation. Thus, it is difficult to accurately describe the critical condition of dynamic recrystallization occurring in the pure copper using a simple relationship between the critical strain \( \varepsilon_c \) and the peak strain \( \varepsilon_p \). Accordingly, in this work, the incremental power balance method of the thermodynamic system was adopted to investigate the critical condition of dynamic recrystallization [15–17]:

\[
\theta = \frac{d\sigma}{d\varepsilon}_{\varepsilon,T}
\]

\[
\frac{d}{d\sigma}\left(-\frac{d\theta}{d\sigma}\right) = 0
\]

where \( \theta \) is the strain hardening rate, and the critical strain \( \varepsilon_c \) is determined through the strain hardening rate \( \theta \) on the flow curve and the inflection point on the flow stress \( \sigma \) curve or minimum value of \( (-d\theta/d\sigma) \) on \( \sigma \) curve. In the actual investigation, a logarithmic relation is mostly used to describe the flow stress \( \sigma \), which can be derived to be:

![Fig. 2 Schematic diagram of thermoplastic deformation stress–strain curve](Image 60x88 to 284x235)
\[ \theta = \sigma \cdot \frac{d \ln \sigma}{d \varepsilon} \tag{4} \]

\[ -\frac{d \theta}{d \sigma} = \left( \frac{d \ln \sigma}{d \varepsilon} \cdot \frac{d (d (d \ln \sigma / d \varepsilon) / d \varepsilon)}{d \ln \sigma / d \varepsilon} \right) \tag{5} \]

\[ \frac{d \ln \theta}{d \varepsilon} = \frac{d \theta}{d \sigma} \tag{6} \]

### 3.3 Dynamic material modeling

Dynamic material modeling (DMM) is established based on the basic principles of large plastic deformation continuum mechanics, physical system simulation and irreversible thermodynamics. By determination of the parameters of the metal plastic deformation under different deformation conditions, features of power dissipation of the materials during plastic deformation are calculated, which can accurately describe the relationship between microstructural evolution of the metal at high temperature deformation and the plastic deformation parameters [18–21]. Under the condition of a given strain \( \varepsilon \) and the deformation temperature \( T \), the strain rate sensitivity exponent \( m \) can be defined by the ratio of the change of the dissipation co-function \( J \) to the dissipation amount function \( G \), as shown in Eq. (7):

\[ m = \frac{dJ}{dG} = \left[ \frac{\varepsilon \ln \sigma}{\varepsilon \ln \dot{\varepsilon}} \right]_{T, \sigma} \tag{7} \]

The dimensionless power dissipation efficiency factor, \( \eta \), can be described by the strain rate sensitivity exponent \( m \), as shown in Eq. (8). To ensure the accuracy of \( m \), a cubic equation function is generally used to fit the relationship between the flow stress \( \ln \sigma \) and \( \ln \dot{\varepsilon} \). Power dissipation efficiency factor, \( \eta \), constitutes the power dissipation map with variations of temperature and strain rate.

\[ \eta = \frac{J}{J_{\max}} = \frac{2m}{m+1} \tag{8} \]

According to the principle of irreversible thermodynamics, dimensionless parameter \( \xi(\dot{\varepsilon}) \) can be used to express the continuous instability criterion at large plastic deformations, as shown in Eq. (9). With the deformation temperature \( T \) and the strain rate \( \dot{\varepsilon} \) as variables, the non-steady flow region with parameter \( \xi(\dot{\varepsilon}) \) smaller than zero constitutes the instability map.

\[ \xi(\dot{\varepsilon}) = \frac{\varepsilon \ln \varepsilon}{\dot{\varepsilon} \ln \dot{\varepsilon}} \frac{m}{m+1} + m < 0 \tag{9} \]

The power dissipation map and the instability map are superimposed to form DMM processing map corresponding to different true strain capacities, \( \varepsilon \), where the shadow area is generally expressed as the instability region, and the digital on the equivalent line represents the power dissipation factor.

### 4 Results and discussion

#### 4.1 Flow stress and thermal activation energy

Equation (10) is usually used to express the effect of the deformation extent on the flow stress [22,23]. It is found that when the strain rate \( \dot{\varepsilon} \) and deformation temperature \( T \) are constant, the flow stress \( \sigma \) of pure copper and strain \( \varepsilon \) conform to the relation of Eq. (11). The calculated stress is in good agreement with the actual stress (Fig. 1).

\[ \sigma = f_1(\dot{\varepsilon}, T) f(\varepsilon) \tag{10} \]

\[ \ln \sigma = a_0 + a_1 \varepsilon + a_2 \ln \varepsilon + a_3 \varepsilon \ln \varepsilon + a_4 \varepsilon^2 \ln \varepsilon + a_5 \varepsilon \ln \varepsilon^2 + a_6 \varepsilon^2 + a_7 \ln \varepsilon^2 + a_8 \varepsilon^3 + a_9 \ln \varepsilon^3 \tag{11} \]

where \( a_0 \sim a_9 \) are constants, which can be determined by curve fitting. The peak stress \( \sigma_p \) and its corresponding peak strain \( \varepsilon_p \) decrease with elevating the deformation temperature and decreasing the strain rate concluded from the analysis of the flow stress curve. This is due to the fact that the recrystallization nucleation takes a longer time at a comparatively low strain rate, and the number of the nucleation increases to make the effect of recrystallization softening stronger than that of the strain hardening. At relatively high deformation temperature, the slip critical resolved shear stress (CRSS) decreases, and the dynamic recovery and dynamic recrystallization lead to reduced dislocation density, which make the peak stress occur in advance at the elevated temperatures. One of the features of the high temperature plastic deformation is that the strain rate is controlled by the thermal activation process. The effect of deformation temperature and strain rate on the flow stress of metal materials can be quantitatively expressed by the following creep equations [24,25]:

\[ A_1 \sigma^n = \dot{\varepsilon} \exp \left( \frac{Q}{RT} \right) = Z \tag{12} \]

\[ A_2 \exp(\beta \sigma) \dot{\varepsilon} = \dot{\varepsilon} \exp \left( \frac{Q}{RT} \right) = Z \tag{13} \]

\[ A'(\sinh(\alpha \sigma))^{n'} \dot{\varepsilon} = \dot{\varepsilon} \exp \left( \frac{Q}{RT} \right) = Z \tag{14} \]

where \( Z \) is the Zener–Hollomon parameter, with the physical meaning of temperature compensated strain rate factors, \( s^{-1} \); \( A_1, A_2, A, n, \beta, \alpha \) and \( n' \) are constants independent of temperature, \( \beta = \alpha \sigma \); \( \sigma \) is the high temperature flow peak stress; \( \dot{\varepsilon} \) is the strain rate, \( s^{-1} \); \( T \) is the deformation temperature, \( K \); \( R \) is the molar gas constant, 8.314 J/(mol·K); \( Q \) is the hot deformation activation energy, which reflects the difficulty level of
hot deformation of materials, kJ/mol. We found that Eqs. (12) and (13) are applicable to low stress deformation and high stress deformation, respectively. Equation (14) is the general form of Eqs. (12) and (13). In the case of $\alpha \sigma < 0.8$, Eq. (14) can be simplified as Eq. (12), while when $\alpha \sigma > 1.2$, Eq. (14) can be simplified as Eq. (13).

When deformation temperature $T$ or strain rate $\varepsilon$ is constant, peak stress $\sigma_p$ exhibits a nearly linear relation with the logarithmic value of strain rate $\dot{\varepsilon}$ or the reciprocal value of temperature, as shown in Fig. 3. It can be seen that the pure copper is in agreement with the Arrhenius equation. For $\beta$ value, the average value of the last three straight slopes in Fig. 3(a) is taken to obtain $\beta = 0.1154$ MPa$^{-1}$. For $n$ value, the average value of the former four straight slopes in Fig. 3(b) is taken to get the $n$ value of 6.7847. Thus, $\alpha = \beta/n = 0.017$ MPa$^{-1}$.

The relationship between $\partial \ln \dot{\varepsilon}$ and $\partial (1/T)$ is described by Eq. (15). At a constant temperature $T$, for $\frac{\partial \ln \dot{\varepsilon}}{\partial \ln[\sinh(\alpha \sigma)]}$, the average slope of seven curves in Fig. 3(c) is obtained to be 5.93904. At constant strain rate $\dot{\varepsilon}$, for $\frac{\partial (1/T)}{\partial \ln[\sinh(\alpha \sigma)]}$, the average slope of four curves in Fig. 3(d) is obtained to be 0.000163. Eventually, the macroscopic activation energy $Q$ is calculated to be 303.8 kJ/mol by Eq. (16).

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

Equation (14) is used to perform the fitting analysis. It is possible to obtain the expression of peak stress $\sigma_p$ as

$$\ln[\sinh(0.017009387 \sigma_p)] = \frac{\ln Z - 34.38327}{6.31637} - 34.38327$$

4.2 Strain hardening and softening

In general, in the calculation of the strain hardening rate $\theta$, the right side of Eq. (2) is approximately equal to $\Delta \sigma / \Delta \varepsilon$, and the fluctuation of the actual test data has a relatively great impact on the calculation accuracy. Herein, the relationship between the flow stress and strain established by Eq. (11) is subjected to derivative processing, as shown in Eqs. (18) and (19). By incorporating them with Eqs. (4) and (5), we can get the strain hardening rate $\theta$ and $-d\theta/d\sigma$. 

![Fig. 3 Relationship between $\ln \dot{\varepsilon}$ and $\sigma_p$ (a), $\ln \dot{\varepsilon}$ and $\ln \sigma_p$ (b), $\ln \dot{\varepsilon}$ and $\ln[\sinh(\alpha \sigma_p)]$ (c) and $T^{-1}$ and $\ln[\sinh(\alpha \sigma_p)]$ (d)](image-url)
\[
\frac{d\ln \sigma}{d\varepsilon} = a_1 + \frac{a_2}{\varepsilon} + a_3(\ln \varepsilon + 1) + a_4(2\varepsilon \ln \varepsilon + \varepsilon^2) + a_5(\ln \varepsilon)^2 + 2a_6\ln \varepsilon + \frac{2a_7}{\varepsilon} + 3a_8\varepsilon^2 + \frac{a_93(\ln \varepsilon)^2}{\varepsilon}
\]

(18)

\[
d\left(\frac{d\ln \sigma}{d\varepsilon}\right) = \frac{-a_2}{\varepsilon^2} + \frac{a_2}{\varepsilon} + a_4(2\ln \varepsilon + 3) + 2a_5\left(\frac{\ln \varepsilon + 1}{\varepsilon}\right) + 2a_6 + 2a_7\left(\frac{1 - \ln \varepsilon}{\varepsilon^2}\right) + 6a_8\varepsilon + 3a_9\left[\frac{2\ln \varepsilon - (\ln \varepsilon)^2}{\varepsilon^2}\right]
\]

(19)

where \(a_1\) to \(a_9\) are coefficients.

Fig. 4 shows the curves of the strain hardening rate \(\theta\) and the strain \(\varepsilon\), from which we can see that the \(\theta\) value decreases gradually as \(\varepsilon\) increases. The strain at \(\theta=0\) is just the peak strain \(\varepsilon_p\). Then, as \(\varepsilon\) increases, \(\theta\) value is gradually restored to 0 again. This exhibits an obvious feature of peak value on the flow stress curve, indicating that a discontinuous dynamic recrystallization occurs in this stage. The strain at which \(\theta\) value returns to 0 for the first time is precisely the steady strain \(\varepsilon_{ss}\). This shows that it is now in the state of full dynamic recrystallization. When further increasing the strain, \(\theta\) value increases again gradually, indicating that the secondary strain hardening is dominant. Accompanying with the increased deformation rate and dropped deformation temperature, the dislocation cross slip and climbing cannot be fully conducted, softening degree provided by the cross slip and climbing is relatively small, rendering the delayed occurrence of peak strain \(\varepsilon_p\).

At a deformation temperature of 600 °C and a strain rate of 1 s\(^{-1}\), \(\theta\) value is larger than 0 all the time, indicating that the strain fails to reach the peak strain \(\varepsilon_p\). In this stage, the strain hardening exceeds dynamic recovery and dynamic recrystallization softening.

The curves of strain hardening rate \(\theta\) and stress \(\sigma\) are shown in Fig. 5, which can be divided into four stages. Stage I was a linear hardening stage, starting from the strain initiation until the commencement of subgrain formation. Consequently, the dynamic recovery rate became slow. The slope of the curve turned to be lowered gradually, and the deformation entered into the second linear hardening stage (Stage II). When the deformation reached the critical strain \(\varepsilon_c\) of a dynamic recrystallization, the strain hardening rate \(\theta\) quickly dropped to 0 in Stage III. After the strain hardening rate \(\theta\)
returned to 0 for the first time, \( \theta \) value increased gradually again with increasing the deformation in Stage IV. Critical stress \( \sigma_c \) could be determined through the inflection point on \( \theta \) vs \( \sigma \) curve or the minimum value on \( (-d\theta/d\varepsilon) \) vs \( \sigma \) curve. The strain corresponding to \( \sigma_c \) value was just the critical strain \( \varepsilon_c \), which became larger with the increase of the deformation rate. The critical strain \( \varepsilon_c \) decreased gradually with the elevation of the deformation temperature. \( \theta \) value was reduced gradually with the increment of the deformation degree. The stress at \( \theta=0 \) was just the peak stress value of \( \sigma_p \). The strain at which \( \theta \) value returned to 0 for the first time was the steady state stress \( \sigma_{ss} \). This result is in line with the analytic result shown in Fig. 4. At a deformation temperature of 800 °C and a strain rate of 0.001 s\(^{-1}\), \( \theta \) value recovered gradually from negative to 0. There was slight difference between the peak stress \( \sigma_p \) and the steady state stress \( \sigma_{ss} \). This result is in line with the analytic result shown in Fig. 4.

Figure 6 show the relationship between the peak strain \( \varepsilon_p \) and the critical strain \( \varepsilon_c \) with the parameter 2. Through regression analysis of the calculated \( \varepsilon_p/\varepsilon_c=0.4-0.6 \), the expression of the peak strain \( \varepsilon_p \) and the critical strain \( \varepsilon_c \) could be obtained, respectively, through Eq. (20) and Eq. (21), being close to the lower limit of empirical formula \( \varepsilon_c/\varepsilon_p=0.6-0.85 \).

\[
\varepsilon_p = 0.000242854Z^{0.10521} \tag{20}
\]

\[
\varepsilon_c = 0.007665772Z^{0.08607} \tag{21}
\]

To calculate the \( \sigma_{rec} \) value of Eq. (1), it is needed to further determine \( \sigma_{sat} \) \( \sigma_0 \) and \( r \) values, where, 2% of the total strain is obtained as the initial yield strain \( \varepsilon_0 \). The initial yield stress \( \sigma_0 \) can be obtained correspondingly. Using Eq. (1), Eq. (22) can be obtained. By incorporating Eqs. (1) and (22) to obtain Eq. (23) to seek derivatives of \( \sigma_{rec} \) and \( \varepsilon \):

\[
\frac{d\sigma_{rec}}{d\varepsilon} = \frac{1}{2} \frac{1}{\sigma_{rec}} - r(\sigma_{sat}^2 - \sigma_{rec}^2) \exp[-r(\varepsilon - \varepsilon_0)] \tag{22}
\]

\[
\frac{d\sigma_{rec}}{d\varepsilon} \sigma_{rec} = \frac{r}{2} (\sigma_{sat}^2 - \sigma_{rec}^2) \tag{23}
\]

When the strain \( \varepsilon \) is smaller than the dynamic recrystallization critical strain \( \varepsilon_c \), \( d\sigma_{rec}/d\varepsilon=\theta \). Thus, Eq. (23) can be derived to

\[
\theta \sigma_{rec} = \frac{r}{2} (\sigma_{sat}^2 - \sigma_{rec}^2) \tag{24}
\]

\[
\frac{d(\theta \sigma_{rec})}{d\sigma_{rec}} = \frac{r}{2} \tag{25}
\]

It can be seen from Fig. 7 that \( \sigma_{rec}^2 \) and \( \theta \sigma_{rec} \) have a linear relationship. Through the linear regression analysis, the slope \( m \) of \( \sigma_{rec}^2 \) vs \( \theta \sigma_{rec} \) curve is obtained. From Eq. (25), \( r=-2m \) is deduced. By extending \( \sigma_{rec}^2 \) and \( \theta \sigma_{rec} \) curves to \( \theta \sigma_{rec}=0 \), the stress \( \sigma_{sat} \) (location indicated by arrow in Fig. 7) can be saturated.

The determined \( \sigma_{sat} \) \( \sigma_0 \) and \( r \) values are substituted into Eq. (1) for computation. Then, the dynamic recovery stress curve \( \sigma_{rec} \) is available, as shown in Fig. 8. Because the effect of the secondary strain hardening is not taken into account in Eq. (1), the \( \sigma_{rec} \) value gradually approaches the saturated stress \( \sigma_{sat} \) when the strain \( \varepsilon \) reaches the peak strain \( \varepsilon_p \).

To further get the dynamic recovery volume fraction \( X_{rec} \) and the dynamic recrystallization volume fraction \( X_{drc} \), it is easily known from Eq. (1) that:

\[
\frac{\sigma_{rec}^2 - \sigma_{sat}^2}{\sigma_{sat}^2 - \sigma_0^2} = 1 - \exp[-r(\varepsilon - \varepsilon_0)] \tag{26}
\]

For a constant strain rate \( \dot{\varepsilon} \), Eq. (26) can be rewritten as

\[
X_{rec} = \frac{\sigma_{rec}^2 - \sigma_0^2}{\sigma_{sat}^2 - \sigma_0^2} = 1 - \exp[-r\dot{\varepsilon}(t-t')] \tag{27}
\]
Fig. 7 Relationship between $\theta$ and $\sigma$ at temperatures of 600 °C (a) and 700 °C (b)

Fig. 8 Comparison between actually measured and calculated $\sigma$−$\varepsilon$ curves at temperatures of 600 °C (a) and 700 °C (b)

\[
\frac{dX_{\text{rec}}}{dt} = r \dot{\varepsilon} (1 - X_{\text{rec}}) \tag{28}
\]

where $t$ is the unloading time, $r$ is the time at which the related phenomenon is initiated ($\tau$ is negligible since DRV starts immediately after straining). At a constant strain rate $\dot{\varepsilon}$, $dX_{\text{rec}}/dt$ increases when $r$ increases. Therefore, the physical meaning of $r$ lies in constructing a bridge between the dynamic recovery kinetics and the deformation conditions.

After the deformation reaches the critical strain $\varepsilon_c$, dynamic recrystallization volume fraction $X_{\text{drx}}$ can be expressed using the Avrami equation [11−13]:

\[
X_{\text{drx}} = 1 - \exp \left[ k \left( \frac{t}{t_{0.5}} \right)^n \right] \tag{29}
\]

where $k$ is the coefficient related to deformation conditions with the value equal to $\ln(1-X_{\text{drx}})$, when the dynamic recrystallization volume fraction is 50%, $k=-0.693$. $n$ is the time exponent, and $n=2$ is taken; $t$ is the unloading time; $t_{0.5}$ is the time when the dynamic recrystallization volume fraction is 50%. As the rates of the nucleation and the grain growth during the dynamic recrystallization are not constant, the crystal nucleus is also not the homogeneous nucleation in the whole volume. According to the theory of recrystallization and model structure [11], we can obtain

\[
X_{\text{drx}} = 1 - \exp \left[ -0.693 \left( \frac{\varepsilon - \varepsilon_c}{\varepsilon_{0.5} - \varepsilon_c} \right)^2 \right] \tag{30}
\]

\[
\varepsilon_{0.5} = \frac{\varepsilon_s + \varepsilon_c}{2} \tag{31}
\]

From Fig. 2, the difference between $\sigma_{\text{rec}}$ and $\sigma_{\text{drx}}$ can be represented as $\Delta \sigma$, which is the net softening directly attributed to DRX. Then, the softening fraction due to DRX can be prescribed by Eq. (32). Meanwhile, after incorporating with Eq. (30), the dynamic recrystallization stress curve $\sigma_{\text{drx}}$ can be achieved:

\[
X_{\text{drx}} = \frac{\Delta \sigma}{\sigma_{\text{sat}} - \sigma_{ss}} = \frac{\sigma_{\text{rec}} - \sigma_{\text{drx}}}{\sigma_{\text{sat}} - \sigma_{ss}} \tag{32}
\]

\[
\sigma_{\text{drx}} = \sigma_{\text{rec}} - (\sigma_{\text{at}} - \sigma_{ss}) \left[ 1 - \exp \left[ -0.693 \left( \frac{\varepsilon - \varepsilon_c}{\varepsilon_{0.5} - \varepsilon_c} \right)^2 \right] \right] \tag{33}
\]
Figure 9(a) shows the dynamic recovery volume fraction $X_{\text{rec}}$ of the material at a deformation temperature of 700 °C. The $X_{\text{rec}}$ value increases sharply as the strain increases. After it reaches the critical strain $\varepsilon_c$, the $X_{\text{rec}}$ value increases slowly. As the strain reaches 0.5, the material is basically in a state of full dynamic recovery. Figure 9(b) shows the dynamic recrystallization volume fraction $X_{\text{drx}}$ of the material at a deformation temperature of 600 °C. The $X_{\text{drx}}$ value increases rapidly after the deformation reaches the critical strain $\varepsilon_c$, and tends to 1 when the strain reaches a steady state strain $\varepsilon_{\text{ss}}$, getting in a state of full dynamic recrystallization. With further increasing strain, the effect of secondary hardening is strengthened, exceeding the dynamic recrystallization softening with a rapidly decreased $X_{\text{drx}}$. With increasing strain rate and decreasing temperature, the dislocation density increases, and in turn higher deformation activation energy is required to drive dynamic recovery and dynamic recrystallization [2,26]. This results in the reduction of $X_{\text{rec}}$ and $X_{\text{drx}}$ values.

4.3 Processing map

Figure 10 shows the processing map of the pure copper at a strain of 0.6. The contour numbers represent the efficiency of dissipation, and the red region corresponds to the flow instability in Fig. 10. It can be seen that there is a large region that corresponds to a feasible processing window, and only one flow instability region can be observed, which occurs in low temperature and high strain rate region (400–450 °C, 0.001–0.05 s$^{-1}$ and 450–750 °C, 0.05–1 s$^{-1}$). Instability mechanisms are likely associated with cracking, localized plastic flow and adiabatic shear bands. Figure 11 shows the optical microstructure of the pure copper in the stability region. It can be seen that many dynamic recrystallization grains can be observed, which suggests that the deformation mechanism in the stability region is dynamic recrystallization.

5 Conclusions

1) With increasing deformation degree, the hot compression deformation of pure copper exhibits features including strain hardening and softening. Before the deformation degree reaches the critical strain, strain hardening and dynamic recovery dominate the deformation, and the flow stress increases significantly. After the deformation degree exceeds the peak strain, the dynamic recrystallization softening enables the flow stress to decrease gradually.

2) The true stress–true strain curves predicted by the
extracted model are in good agreement with the experimental results. The expression of strain hardening rate, dynamic recrystallization critical stress, saturated stress, dynamic recovery volume fraction and dynamic recrystallization volume fraction are determined.

3) The peak value feature on the flow stress curve is unable to uniquely characterize the dynamic recrystallization softening behavior. The expressions of the dynamic recrystallization critical strain and the peak strain that can be expressed as \( \varepsilon_c/\varepsilon_p = 0.4 \sim 0.6 \).

4) According to the processing map established, the instability region occurs in regions of 400–450 °C, 0.001–0.05 s\(^{-1}\) and 450–750 °C, 0.05–1 s\(^{-1}\). The deformation mechanism in the instability region is dynamic recrystallization.

References


应变速率和变形温度对纯铜应变硬化和软化行为的影响

黄树海，舒大禹，胡传凯，朱世凤

西南技术工程研究所，重庆 400039

摘 要：通过热压缩实验研究变形温度和应变速率对纯铜热变形行为的影响，确定了应变硬化率、动态再结晶临界应力、饱和应力、动态回复体积分数和动态再结晶体积分数的表达式。结合热加工图，得到纯铜的失稳区域主要位于400~450 °C、0.001~0.05 s⁻¹和450~750 °C、0.05~1 s⁻¹区间，稳态区域的变形机制主要为动态再结晶。对流变应力进行预测，预测结果与实验结果吻合较好。

关键词：铜；热压缩变形；应变硬化；应变软化

(Edited by Wei-ping CHEN)