Materials Research Express

PAPER

Flow stress modeling, processing maps and microstructure evolution of 05Cr17Ni4Cu4Nb Martensitic stainless steel during hot plastic deformation

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Keywords: Microstructure, flow stress modeling, hot processing maps, 05Cr17Ni4Cu4Nb martensitic stainless steel

Abstract
The hot deformation behavior of 05Cr17Ni4Cu4Nb stainless steel at the temperatures from 950 °C to 1250 °C and strain rates from 0.01 s⁻¹ to 5 s⁻¹ was investigated on the basis of test data obtained by Gleeble1500D thermo-simulation machine. A two-stage flow stress constitutive model incorporating the effects of the strain rate and temperature on the deformation behavior is proposed. The stress-strain relations of 05Cr17Ni4Cu4Nb steel predicted by the proposed model agree well with the experimental results. Moreover, the hot processing maps are also investigated. Combined with the microstructure evolution analysis, the appropriate hot forming processing parameters of 05Cr17Ni4Cu4Nb steel is proposed in the temperature range of 1000–1100 °C and strain rate range of 0.01–0.02 s⁻¹.

1. Introduction

With the characteristics of anti-corrosion, resist-to-abrasion and high mechanical strength, 05Cr17Ni4Cu4Nb martensitic stainless steel is widely used in structural components of the equipment in marine construction, chemical industries and power plants with a year-to-year consumption increase [1]. Recently, many studies about 05Cr17Ni4Cu4Nb martensitic stainless steel focus on its heat treatment processes, surface coating methods or the properties during additive manufacturing [2–5]. Generally, hot forging process is the effective method to form the shape and refine the grains for the parts made of 05Cr17Ni4Cu4Nb steel. To improve the mechanical performances of this material, the designing for the deformation processing parameters of hot forging should be investigated, which requires the further researches on the material flow behaviors, processing maps and the microstructure evolution.

To improve the forming parameters and develop the reliability of hot forging process, finite element (FE) simulation can be effective method to promote the shape-controlling and property-controlling forging technology. To carry out an effective evaluation of the material flow behavior and the stress/strain distribution during hot forging process, it is essential to establish the precise constitutive equation that describes the combined influences of temperature, strain rate, and strain on the flow stress. Generally, two kinds of common methods are adopted in establishing the flow stress constitutive model: phenomenological method and physical-based method, which have been applied for different types of materials [6–10]. The phenomenological models including Johnson–Cook (JC) model [11], Arrhenius equation [12], etc are mainly based on the mathematical equations in order to fit or regress the experimental data, therefore, most of the parameters in phenomenological models lack physical meanings. During the hot deformation process, the internal microstructure of metals generally evolves extensively and significantly influences the constitutive behavior of materials [13]. Investigations of the hot deformation behaviors for different martensitic steels(30Cr15Mo1N [14], AISI403/403Nb [15], high nitrogen martensitic steel [16], etc) have been carried out carefully, and it was found that the dynamic recrystallization (DRX) occur during the hot deformation conditions. However, the constitutive
models in the literatures were established on the basis of Arrhenius equation, which were not quite suitable to
give a reasonable description on the mechanism of DRX. To solve this problem, the physical-based models are
established on the framework of deformation mechanism, which take into account the evolution of dislocation
density and recrystallization kinetics. Generally, the true stress-true strain relations of steels in hot
deformation exhibit the characteristics of working hardening, recovery and DRX obviously. Therefore, the
physical-based two-stage constitutive model was proposed to describe the hardening, softening and steady
behavior in different stages of the true stress-true strain curves. The researches in literatures exhibit the good
applicability and higher accuracy of the two-stage physical-based constitutive model for various kinds of alloys,
including the Ni-based superalloys, steels, etc. In addition, the hot flow stress behavior of martensitic steels(AISI 410, 2Cr11Mo1VNbN, FB2 rotor steel, etc) were described by the two-stage physical-based model in the literatures, and the characteristics of the flow stress were captured. Although the
physical-based model has been used in several kinds of alloys, researches are still lack in analyzing the hot
deformation behavior of 05Cr17Ni4Cu4Nb martensitic stainless steel. In the presented work, the two-stage
physical-based model is adopted to establish the stress-strain relations of this steel, considering the experimental
flow stress curves of this steel exhibit the typical DRX feature. The DRX fraction is implied in the presented two-stage
model, which is related to the peak strain $\varepsilon_p$ and critical strain $\varepsilon_c$ and determined from the data of saturated
stress $\sigma_{sat}$ and steady state stress $\sigma_0$ obtained from the experimental flow stress curves. Considering the variables
in the equations represent corresponding physical meanings, the presented two-stage model gives a more
reasonable interpretation to the constitutive properties of 05Cr17Ni4Cu4Nb steel including the dependences of
the deformation temperature, strain rate and strain history.

The optimization of the plastic deformation conditions of steels is gaining increasing attention in the
research and development of hot forming technologies. Based on the dynamic materials model(DMM), the
processing map technology has been established as an effective approach to investigate the deformation

![Figure 1](image1.png)

Figure 1. (a) Scheme of the hot compression test and (b) the initial microstructure of the studied steel before deformation.

![Figure 2](image2.png)

Figure 2. (a) Scheme of the hot compression test. (b) Typical stress-strain curve of steels in hot forming. Typical true stress-strain
curves of 05Cr17Ni4Cu4Nb at the deformation condition: (c) 1250 °C and (d) 0.1 s$^{-1}$. 
mechanisms of materials and find the optimal processing parameters. It has been applied in various kinds of materials including steels [30, 31], alloys of aluminum, magnesium, etc. Xu et al [32] determined that the optimum condition for 25Cr3Mo3NiNb steel was at 1077–1177 °C and 0.001–0.03 s −1. Cai et al [33] suggested 1120 °C–1160 °C and 0.03–0.1 s −1 as the optimum process parameters by analyzing the processing maps. As one kind of the most widely used materials, stainless steels are studied by many researchers, and the hot deformation mechanism and optimum processing parameters are revealed according to the processing maps [34–38]. Ren et al [39] researched the hot deformation characteristics of AISI 420 stainless steel, and specified that the hot deformation should be carried out at the temperature conditions of 1007–1087 °C and the strain

![Typical stress-strain curve of the steel in hot forming.](image)

**Figure 3.** Typical stress-strain curve of the steel in hot forming.

![Relationships between (a) ln \( \dot{\varepsilon} \) and ln \( \sigma \), (b) ln \( \dot{\varepsilon} \) and \( \sigma \), (c) ln \( \dot{\varepsilon} \) and ln \( \left[ \sinh \left( \sigma \right) \right] \), (d) ln \( \left[ \sinh \left( \sigma \right) \right] \) and 10000/\( T \).](image)

**Figure 4.** Relationships between (a) ln \( \dot{\varepsilon} \) and ln \( \sigma \), (b) ln \( \dot{\varepsilon} \) and \( \sigma \), (c) ln \( \dot{\varepsilon} \) and ln \( \left[ \sinh \left( \sigma \right) \right] \), (d) ln \( \left[ \sinh \left( \sigma \right) \right] \) and 10000/\( T \).
rate of 0.01–0.05 s$^{-1}$ when the strain is no less than 0.5. During the hot forming process, deformation defects caused by the instable plastic flow of the materials can be reduced and avoided on the basis of the investigation of the processing maps.

In this paper, the effects of thermo-mechanical parameters on the hot deformation behavior of 05Cr17Ni4Cu4Nb martensitic stainless steel are investigated. Based on the experimental data, a two-stage flow stress constitutive equations have been established. By using the processing map approach associated with the microstructure investigation, the appropriate processing conditions for this material are determined.

2. Experiment material and methods

The uniaxial isothermal compression tests of 05Cr17Ni4Cu4Nb steel were performed by using the Gleeble1500D thermo-mechanical simulator. $\varphi$ 8 mm $\times$ 12 mm cylindrical specimens for the tests were manufactured. The experiments procedures are illustrated in figure 1(a). The compression tests were carried out under various temperatures (950 °C, 1050 °C, 1150 °C, 1250 °C) and strain rates (0.01 s$^{-1}$, 0.1 s$^{-1}$, 1 s$^{-1}$, 5 s$^{-1}$) conditions. During the hot deformation test, the slices of graphite were used between the specimen and anvil to reduce the friction. The deformed specimens were immediately quenched in water after the hot compressive deformation with a true stain of 0.8. Then, the specimens were split and the axial sections were polished and etched in boiled aqueous solution of sulphuric acid with potassium permanganate to capture the micrographs (sulphuric acid 10% by volume 100 g + potassium permanganate 1 g). The optical and SEM observations are carried out. SEM study was carried out on the Shimadzu SS550-SEM scanning electron microscope with the operation voltage 15 kV. Figure 1(b) shows the initial microstructure of the steel before hot deformation. The equiaxed grains with the average size of 26.4 $\mu$m are observed.
3. Results and discussion

3.1. Flow features and hot deformation mechanism

In the compressive tests, the friction and heat of deformation may influence the flow stress data. According to the measurements of different specimens, the values of the barreling coefficient $B = \frac{h R_M^2}{h_0 R_0^2}$ reported by Roebuck [40] are calculated. $h_0$ and $h$ are the heights of the specimen before and after deformation, respectively. $R_0$ and $R_M$ are the initial radius before deformation and the maximum radius after deformation of the specimen. In this study, the values of $B$ for different specimens are no more than 1.1, thus, the experimental flow stress curves are valid and the effect of friction on the flow stress is ignored. Besides, the heat deformation effect was removed by using the interpolation method according to the fitted relationship between $\sigma$ and $1/T$.

Figure 2 shows the true stress–true strain curves of 05Cr17Ni4Cu4Nb steel at the temperature range of 950 °C–1250 °C and the strain rate range of 0.01 s$^{-1}$–5s$^{-1}$. The flow stress curves exhibit the combined effects $\ln \sigma$, $\ln \varepsilon$, and $\ln Z$. The relationships between $\ln \sigma$, $\ln \varepsilon$, $\ln Z$, $\ln [\sinh (\alpha \varepsilon)]$, and $\ln [\sinh (\alpha \sigma)]$ are shown in Figure 6.
of the strain, strain rate and temperature. In the initial stage of deformation, the flow stress rises rapidly with the increasing of strain due to the work hardening (WH) caused by dislocation generation and multiplication [41, 42]. With the increasing of strain, the effect of WH is offset partially by dynamic softening, including dynamic recovery (DRV) and dynamic recrystallization (DRX). After reaching the peak value, the flow stress gradually decreases to a relatively steady state due to the equilibrium achieved between WH and dynamic softening. Figure 3 demonstrates the two typical flow features of metals in hot working process. As shown as the dash line, if DRX scarcely occurs, the flow stress increasingly rises from the initial stress $s_0$ to the saturated stress $s_{sat}$ due to the effect of WH and DRV. Correspondingly, as shown as the solid line, the trend of flow stress is in accordance with the dash line in the first stage of strain $\varepsilon < \varepsilon_c$ (where $\varepsilon_c$ is the critical strain). With the increasing of strain, the accumulated dislocation density increases and then exceeds the critical value to induce the DRX occurring. Due to the combined effects of WH, DRV and DRX, the flow stress first increases to the peak stress $s_p$ and then descends and stabilizes at the steady state stress $s_{ss}$ during the second stage ($\varepsilon \geq \varepsilon_c$).

As shown in figure 2, the flow stress curves of the hot compression test in this study are similar to mode of the solid line, therefore a physical-based two-stage equation is applied to establish the constitutive model of this steel.

### 3.2. Establishment of Constitutive relations

According to presented researches, the evolvement of dislocation density $\rho$ can be represented as [23–25, 28, 43]:

$$\frac{d\rho}{d\varepsilon} = k_1 \sqrt{\rho} - k_2 \rho$$

(1)

where $k_1$ and $k_2$ are the coefficients related to working hardening and dynamic recovery. When the strain $\varepsilon$ is zero, $\rho$ equals to the initial dislocation density $\rho_0$. According to equation (1), the dislocation density $\rho$ can be expressed as follows:

$$\rho = \left( \frac{k_1}{k_2} - \frac{k_1}{k_2} e^{-\frac{k_2}{k_1}} + \sqrt{\rho_0 e^{-\frac{k_2}{k_1}}} \right)^2$$

(2)

when $d\rho/d\varepsilon = 0$, $\rho_s = (k_1/k_2)^2$, where $\rho_s$ is the saturation dislocation density which is corresponding to $\sigma_{sat}$. Due to the effective stress is negligible compared to the internal stress at high temperature, the applied stress can be approximately calculated by $\sigma = \gamma \mu b \sqrt{\rho}$, where $\gamma$ is the material constant, $\mu$ is the shear modulus, and $b$ represents the Burgers vector. Then, the stress $\sigma_{WH}$ related to the strain $\varepsilon$ at the working hardening and dynamic recovery stage can be described as follows:

$$\sigma_{WH} = \sigma_{sat} + (\sigma_0 - \sigma_{sat}) e^{\frac{k_2}{k_1} \varepsilon}, \quad (0 < \varepsilon < \varepsilon_c)$$

(3)

The dynamic recrystallization during the deformation process occurs when the strain $\varepsilon$ reaches the critical strain $\varepsilon_c$, thus the effect of DRX on the flow stress should be considered. The dynamic recrystallization volume fraction $X_{drx}$ can be represented as the following equation [44, 45]:

![Figure 7. Relationships between $\ln(-\ln(1-X_{drx}))$ and $\ln(\varepsilon - \varepsilon_c)/\varepsilon_p$](image)
where \( \varepsilon_s \) is the strain corresponding to peak stress \( \sigma_p \), \( k_{\text{dRX}} \) and \( n_{\text{dRX}} \) are material constants. As shown in figure 3, the descent of the stresses between solid line and dash line reflects the effect of DRX, thus the relationship between \( X_{\text{dRX}} \) and the stress parameters can be represented as [46]:

\[
X_{\text{dRX}} = \frac{\sigma_{\text{WH}} - \sigma}{\sigma_{\text{sat}} - \sigma_{\text{sat}}}, \quad (\varepsilon \geq \varepsilon_s)
\]  

According to equations (4) and (5), the constitutive equation can be derived as:

\[
\sigma = \sigma_{\text{WH}} - (\sigma_{\text{sat}} - \sigma_{\text{sat}}) \left( 1 - \exp \left( -k_{\text{dRX}} \left( \frac{\varepsilon - \varepsilon_s}{\varepsilon_p} \right)^{n_{\text{dRX}}} \right) \right), \quad (\varepsilon \geq \varepsilon_s)
\]  

In the above description, the framework of the flow stress model for 05Cr17Ni4Cu4Nb steel has been established. Generally, the relations among strain rate \( \dot{\varepsilon} (\text{s}^{-1}) \), temperature \( T (\text{K}) \), activation energy \( Q (\text{J/mol}) \) and Zener-Holloman parameter can be described in different ways as:

\[
\dot{\varepsilon} = A_1 \sigma^{n_1}
\]  
\[
\dot{\varepsilon} = A_2 \exp (\beta \sigma)
\]  
\[
\dot{\varepsilon} = A \exp (Q/RT) = A (\sinh (\alpha \sigma))^\alpha
\]

where \( A_1, A_2, A, n_1, n, \beta \) and \( \alpha \) are constants need to be determined and \( \alpha = \beta / n_1 [47, 48] \). 

\( R = 8.314 \text{ J/(mol·K)} \). The logarithm of equations (7), (8) and (9) are calculated respectively, and the relations are specified as:

\[
\ln \dot{\varepsilon} = \ln A_1 + n_1 \ln \sigma - (Q/RT)
\]  
\[
\ln \dot{\varepsilon} = \ln A_2 + \beta \sigma - (Q/RT)
\]  
\[
\ln \dot{\varepsilon} = \ln A + \alpha \ln (\sinh (\alpha \sigma)) - (Q/RT)
\]  

Figure 8. The calculated curves and the experimental data.
According to the flow stress curves in this investigation, $\sigma_r$ are used to regress and determine the values of $A$, $n_r$, $n_1$, $\beta$, $\alpha$ and $Q$ as shown in figure 4. Based on equations (10) and (11), the mean values of $n_1$ and $\beta$ (the average slopes of lines in figures 4(a) and (b)) are fitted as 6.604 and 0.07 MPa$^{-1}$, respectively. This determines the value of $\alpha = \beta / n_1 = 0.0106$ MPa$^{-1}$. According to equation (7), the values of $n$ and $\ln A$ is associated with the mean slope and intercept of $\ln \hat{\eta}$ against $\ln (\sinh (\alpha \sigma))$ in figure 4(c). Thus, the mean values of $n$ and $A$ are determined as 5.251 and 7.21 $\times$ 10$^{14}$ s$^{-1}$. Then, the partial differentiation of equation (12) is:

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

(13)

where $Q$ is related to the slope of the plot for $\ln (\sinh (\alpha \sigma))$ against $1/T$ (figure 4(d)). Based on the calculated data, $Q$ is determined as 426822.1 J/mol. Generally, the activation energy of hot deformation is considered to be influenced by various factors such as element composition, initial microstructure, experimental conditions, etc. It is found that the higher deformation activation energy often appears in the hot deformation of alloys with higher yield strength [49]. In literatures, the values of $Q$ are determined as 376.5 kJ/mol [50] and 892.35 kJ/mol for 34CrNiMo6 steel and 3Cr20Ni10W2 steel [49], respectively. With the rising of the alloying elements fraction,
it indicates that the deformation activation energy tends to enlarge obviously. The alloying elements enhance the strength of steel by increasing the obstruction of the dislocation moment and grain boundary migration, and thus lead to the increment of the hot deformation activation energy.

In hot plastic deformation, the working hardening rate is defined as:

\[ \theta = \frac{\partial \sigma}{\partial \varepsilon} \bigg|_{\varepsilon, T} \]  

Figure 5 shows the \( \theta - \sigma \) curves under different deformation conditions. When \( \theta = 0 \), \( \sigma \) are corresponding to \( \sigma_p \) and \( \sigma_{ps} \), and the value of strain which relates to \( \sigma_p \) in the flow stress curves is the peak strain \( \varepsilon_p \). According to the test data, the relationship between \( \ln \varepsilon_p \) and \( \ln Z \) is illustrated in figure 6(a), and the mathematical model of \( \varepsilon_p \) can be determined as:

\[ \varepsilon_p = 0.002973Z^{0.12154} \]  

Using the similar method, \( \sigma_{ps} \) can be obtained as shown in figure 6(c) and modeled as:

\[ \sigma_{ps} = 94.75125 \sinh^{-1}(0.000795Z^{0.2007}) \]  

According to figure 5, \( \theta \) is fitted as a cubic function of \( \sigma \) in the range of \( \theta > 0 \). When \( \sigma = \sigma_c \), the second derivative \( \theta'' = \frac{\partial^2 \theta}{\partial \sigma^2} = 0 \). Thus, \( \sigma_c \) can also be determined. Then, according to the experimental flow stress data, \( \varepsilon_c \), which is corresponding to \( \sigma_c \), can be obtained. In figure 6(b), the relationship between \( \varepsilon_c \) and \( Z \) is established approximately as:

\[ \varepsilon_c = 0.002648Z^{0.12154} \]  

The saturated stress \( \sigma_{sat} \) can not be determined in the \( \theta - \sigma \) curves directly. As shown in figure 5, according to the tangent line drawn at \( (\sigma_c, \theta_c) \), the values of \( \sigma_{sat} \) can be obtained and shown in figure 6(d), which is modeled as:

\[ \sigma_{sat} = 94.75125 \sinh^{-1}(0.000738Z^{0.2093}) \]  

The initial stress \( \sigma_0 \) can be determined directly from the experimental data, and the relationship between \( \sigma_0 \) and \( Z \) is shown in figure 6(e). The following expression can be specified:
The constant $k_2$ is calculated by equation (20). Based on the data as shown in figure 6(f), $k_2$ can be expressed by equation (21):

$$k_2 = \ln \left( \frac{\sigma - \sigma_c}{\sigma_0 - \sigma_c} \left( -\frac{2}{\varepsilon} \right) \right)$$

$$k_2 = 157.34 Z^{-0.05123}$$

According to equation (4), $X_{drc}$ can also be given as follows:

$$\ln \left( - \ln \left( 1 - X_{drc} \right) \right) = \ln k_{drc} + n_{drc} \ln \left( \frac{\varepsilon - \varepsilon_c}{\varepsilon_p} \right), \ (\varepsilon \geq \varepsilon_c)$$

Based on equation (5), $X_{drc}$ is calculated from the experimental flow stress data. The relationship between $\ln \left( - \ln \left( 1 - X_{drc} \right) \right)$ and $\ln((\varepsilon - \varepsilon_c)/\varepsilon_p)$ is shown in figure 7. According to the slope and intercept of the fitted line, $k_{drc} = 1.2406$ and $n_{drc} = 0.4698$ are obtained.
According to the above analysis, the parameters in the proposed two-stage constitutive equation of 05Cr17Ni4Cu4Nb steel have been determined on the basis of the hot compression test data. According to this model, the flow stress at high temperatures can be calculated and predicted. Figure 8 shows the comparison between the predicted results of this model and the experimental data of hot compression tests. The predicted flow stress values are in good accordance with the experimental data. To further verify the applicability of the proposed constitutive model, the correlation coefficient $R$ and the average absolute relative error $AARE$ are adopted, which are expressed as equations (23) and (24), respectively:

\[ R = \frac{\sum_{i=1}^{N} (E_i - \bar{E})(P_i - \bar{P})}{\sqrt{\sum_{i=1}^{N} (E_i - \bar{E})^2 \sqrt{\sum_{i=1}^{N} (P_i - \bar{P})^2}}} \]  
\[ AARE = \frac{1}{N} \sum_{i=1}^{N} \left| \frac{E_i - P_i}{E_i} \right| \times 100\% \]
where $E_i$ and $P_i$ are the experimental data and the predicted values, respectively. $N$ is the amount of data used in the investigation. According to equations (23) and (24), the values of $R$ and $AARE$ for the presented constitutive equation are 0.994 and 5.29%, respectively, which indicate that the proposed model can give a high prediction accuracy for the flow stress of this steel. The steels generally obey the von Mises yield criterion and can be treated as the isotropic materials during the plastic deformation at high temperature, therefore, the proposed two-stage constitutive model is available to apply in the numerical simulations of different kinds of hot forging processes and provide credible predictions of the flow behavior of 05Cr17Ni4Cu4Nb steel.

3.3. Processing maps in hot deformation

The processing map is a widely used approach to optimize the processing parameters during hot forging. The processing map method considers the workpiece as a power dissipater during the plastic deformation. The total dissipated power $P$ absorbed by the workpiece can be divided into two parts, the content $G$ for temperature rise and the co-content $J$ for microstructure dissipation. The total dissipated power can be described as:

$$ P = G + J = \int_0^\varepsilon \sigma d\varepsilon + \int_0^\sigma \varepsilon d\sigma $$

(25)

In the meanwhile, the relationship between stress and strain rate is considered to obey the power law:

$$ \sigma = K\dot{\varepsilon}^m $$

(26)

$K$ is the material constant. The strain rate sensitivity $m$ is represented as:

$$ m = \frac{\partial f}{\partial G} = \frac{\dot{\varepsilon} d\sigma}{\sigma d\dot{\varepsilon}} = \frac{d\ln\sigma}{d\ln\dot{\varepsilon}} $$

(27)

The ideal power dissipation $J_{\text{max}}$ occurs when $m$ equals 1.0 in equation (26). The parameter $\eta$, which represents the power dissipation efficiency, is calculated by:

$$ \eta = \frac{J}{J_{\text{max}}} = \frac{J}{\sigma\dot{\varepsilon}/2} = \frac{2m}{m + 1} $$

(28)

The instability parameter $\xi$ can be expressed by using the following relation [28]:

$$ \xi(\dot{\varepsilon}) = \frac{\partial \ln (m/(m + 1))}{\partial \ln \dot{\varepsilon}} + m < 0 $$

(29)

The negative value of instability parameter $\xi$ indicates the occurrence of unstable plastic deformation. In this paper, $\eta$ and $\xi$ are calculated and demonstrated as the processing maps for 05Cr17Ni4Cu4Nb steel in hot forming process, as shown in figure 9. In figures 9(a)–(d), the contour line represents the value of $\eta$ and the shaded region represents the instability conditions in which $\xi < 0$. The three-dimensional isosurfaces of processing maps are demonstrated in figures 9(e) and (f).

According to the contour lines in figures 9(a)–(d), the distributions of $\eta$ are changed with the increasing of strain. When strain reaches 0.4, the peak value of $\eta$ occurs at two regions. One is at (1050 °C, 0.01 s⁻¹), while the other is at (1200 °C, 5 s⁻¹). The high value of $\eta$ indicates that more energy has been dissipated for microstructure evolution (such as DRX) and induces relatively fine grains. In the meanwhile, considering the occurrence possibility of grain coarsening at high temperature and large resistance to plastic flow at large strain rate, the appropriate parameters of hot forming can be selected from the region near the first peak value condition (1050 °C, 0.01 s⁻¹) of $\eta$ when the strain is more than 0.4.

According to shade region shown in figures 9(a)–(d), it indicates that the change of strain also influences the distribution of instability parameter $\xi$. With the increasing of strain, the region of $\xi < 0$, in which the instable flow may occur, firstly shrinks obviously and then fluctuates slightly. The instability area is relatively small at the strain of 0.4 (figure 9(b)). When $\varepsilon > 0.4$, the area of instability region is demonstrated at the temperature 950–1060 °C and the strain rate 0.22–5 s⁻¹.

3.4. Microstructure evolution in hot deformation

Microstructures of 05Cr17Ni4Cu4Nb specimens compressed to the true strain 0.8 were observed by optical microscope and SEM. The evolution of microstructures under different deformations is aggregated in figure 10. The microstructure is significantly influence by deformation temperature and strain rate, and the grain morphology shows the typical DRX occurs. Figure 11(a) shows the grain diameters of the steel after the deformation at different conditions, which demonstrates that the average grain size enlarges with the rising of $T$ and the decreasing of $\dot{\varepsilon}$. Figure 11(b), (c) and (d) demonstrated the microstructures of 05Cr17Ni4Cu4Nb steel under the deformation conditions of (950 °C, 5 s⁻¹), (1050 °C, 5 s⁻¹), and (1050 °C, 1 s⁻¹). It shows that intergranular cracks and voids appear and may lead to the instable deformation, which verify the rationality of the predicted instability region in figure 9(d). Figure 11(e) demonstrated the microstructure of this steel under
the condition of 1150 °C and strain rate 5 s⁻¹, with η = 0.3. The fine grains with necklace shape are observed, which indicate the DRX occurred during the deformation. However, there is no enough time for the migration of the fine DRX grain boundaries during the deformation with high strain rate. Thus, the uneven microstructure composed of the large grown grains and fine DRX grains is obtained under this condition, which leads to the non-uniform mechanical property of the material. Figure 11(f) shows the microstructure of the steel under the condition of 1250 °C and strain rate 5 s⁻¹, with η = 0.36. Compared with figure 11(e), the grain size enlarges with the increasing of the temperature. Moreover, it can be seen that δ-ferrite appears in the matrix of the steel and is deformed severely in figure 11(f), which also causes the non-uniform deformation and stress concentration. Therefore, although the relative high values of η are obtained, these deformation conditions with the microstructures as shown in figures 11(e) and (f) are not the appropriate processing conditions. Figure 12(a) shows the microstructure in the deformation condition of 1050 °C and 0.01 s⁻¹, with η = 0.43. It can be seen that the equiaxed grains and uniform microstructure are obtained. As shown in figures 12(b) and (d), the homogeneity of the microstructure deteriorates with the increasing of strain rate. In figures 12(c), (e) and (f), the grain is coarsening rapidly in the deformation process when the strain rate is relative low and the temperature reaches 1150 °C, as the high temperature and long maintaining time promote the migration of grain boundary. Moreover, the fraction of δ-ferrite separated out from the matrix is prone to increase during the deformation process with relative high temperature and low strain rate. Considering the δ-ferrite may cause the initialization of micro-cracks, deformatinos at high temperature (more than 1150 °C) with long maintaining time should be avoided in the hot forming process of this steel.

4. Conclusions

The hot deformation characteristics of 05Cr17Ni4Cu4Nb steel have been studied and the results are summarized as follows:

(1) The flow stress model of 05Cr17Ni4Cu4Nb steel is proposed with a two-stage form. This model can capture the features of flow stress curves reasonably and provide a good prediction of the flow stress, and can be applied for numerical simulation of the thermal plastic processing.

(2) The hot processing map of this steel is constructed. It indicates that when the strain is larger than 0.4, the hot forming processes are preferred to be carried out at 1000–1100°C and 0.01–0.02 s⁻¹. After the hot forming process in the condition (1050 °C 0.01 s⁻¹) with the peak value of η = 0.43, fine and homogenous grains are obtained in the steel. During the deformation under the conditions (relative low deformation temperature and high strain rate) which locate in the instable region of the hot processing map, intergranular cracks and voids appear in the microstructures of the steel.

(3) Considering the precipitation of δ-ferrite and the coarsening of grains, the deformation under high temperature with long maintaining time should be avoided in the hot forming process.

Acknowledgments

The authors gratefully acknowledge the financial support from National Natural Science Foundation of China (Grant No. 51805204), Postdoctoral Science Foundation of China (Grant No. 2017M621208) and Education Department of Jilin Province in China (Grant No. JJKH20200976KJ).

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