Hot Deformation Behavior and Dynamic Recrystallization Characteristics in a Low-Alloy High-Strength Ni–Cr–Mo–V Steel

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Abstract

This study presented a quantitative investigation of deformation behavior and dynamic recrystallization of low-alloy high-strength Ni–Cr–Mo–V steels during hot deformation. A series of isothermal compression experiments were performed at temperatures ranging from 800 to 1200 °C and strain rates from 0.01 to 10 s^{−1} with a height reduction of 60%. A complete Arrhenius constitutive model and processing maps were developed. The results showed that the constitutive model had the ability to predict the flow stress with an average absolute relative error of < 5.7%. The processing maps constructed at strains of 0.2, 0.4, and 0.8 showed that flow instability was prone to occur at higher strain. Dynamic recrystallization tended to take place at higher temperatures (900–1200 °C) and lower strain rates (0.01–1 s^{−1}). The critical strain for the onset of dynamic recrystallization was determined, and a kinetics model was developed. The predicted results for recrystallization volume fraction and flow stress were compared with the experimental data, which indicated that the model was accurate and reliable.

Keywords Deformation behavior · Dynamic recrystallization · Constitutive equation · Processing map · Critical strain · Recrystallization kinetics

1 Introduction

In recent years, low-alloy high-strength steels have been widely used in the fields of ship building, the drilling platforms of offshore oil and gas, and aerospace manufacture, due to their mechanical properties of high strength with good toughness and ductility without the occurrence of fracture and failure [1]. In these fields, many requirements are necessary to ensure the proper microstructures, excellent properties, and good dimensional stability [2]. To satisfy these requirements, the materials have to be subjected to complicated hot deformation and microstructural

evolution. On the one hand, a given hot deformation processing route yields a particular metallurgical phenomenon (microstructural evolution); on the other hand, the microstructural evolution during the hot deformation process, in turn, influences the mechanical properties of the material [3]. Thus, it is significant for designers of material-forming processes to understand hot deformation mechanisms and microstructural evolution of materials [4–5].

Hot deformation is an effective method for shaping metals. To precisely simulate the deformation process, many constitutive models have been proposed. These can be divided into phenomenological, physical-based, and artificial neural network models [6]. Among these, the Arrhenius equation is most widely used to describe the relationships between the strain rate, flow stress, and temperature, especially at high temperatures [7]. It can accurately predict nonlinear histories under different conditions.

Mandal et al. [8] developed constitutive equation to predict the flow stress of a Ti-modified austenitic stainless steel. Cai et al. [9] developed constitutive equations for different phase regimes to study the workability of Ti6Al4 V alloy. Many other researchers developed artificial neural network models to predict flow stress during hot deformation [6].

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Optimizing the processing parameters for the materials is also important. A processing map based on the dynamic material models is an effective method to control and optimize the processing parameters. Processing map technology was first proposed by Frost and Ashby [10] to optimize hot-working processing parameters and to control the microstructural evolution in the hot deformation of plain low-carbon steel. Subsequently, Prasad et al. [11] and Murty and Rao [12] made further modifications for this technology. In recent years, many researchers have successfully applied processing map technology to optimize the processing parameters of various metals, such as high-strength steels [13], austenitic stainless steels [14], Mg alloys [15], and Al alloys [16].

During a thermo-mechanical process, some microstructural evolutions, such as grain coarsening, phase transformation, precipitation, and recrystallization, can take place, all of which have an important effect on the processing parameters and mechanical properties of parts. Dynamic recrystallization (DRX) is a typical thermo-mechanical process that occurs in metals with medium or low stacking fault energy under certain deformation conditions. DRX can replace deformed microstructures with coarse or elongated grains by newly equiaxed recrystallized structures, which gives the material good mechanical properties [17]. Thus, it is important for engineers to control a microstructure by optimizing the processing parameters for promoting DRX. However, DRX is very sensitive to the processing parameters. It is necessary to clarify the DRX kinetics and microstructural evolution, which are beneficial to optimizing the processing parameters and controlling the microstructure. Many authors have therefore paid much attention to this subject. Chen et al. [18] investigated the DRX characteristics of 30Cr2Ni4MoV ultra-super-critical rotor steel during hot deformation. Najafizadeh and Jonas [19] proposed a theory to determine the critical conditions for DRX. Chen et al. [20] studied the DRX of superalloys by hot compression tests. They estimated the recrystallized volume fraction and grain size based on DRX models, and the results were compared with experimental values. Wei et al. [21], Liu et al. [22], and Zhu et al. [23] investigated the DRX behaviors of various metallic materials. Despite the large efforts invested in studies of the deformation behavior and DRX of various metallic materials, there is, however, little information available concerning the deformation behavior and DRX of the 10Ni5CrMoV steel. An investigation of this subject is therefore of necessity.

In this study, the deformation behavior of the 10Ni5CrMoV steel was experimentally studied and an Arrhenius-type whole-process constitutive equation was developed. To validate the model, the predicted results were compared with the experimental data. To further study the deformation behavior, processing maps at

different strains were established to explain the deformation mechanism. DRX of the material was studied using optical observation. The critical condition for the onset of DRX was determined, and its kinetics model was developed. To validate the model, the predicted results for DRX volume fraction and flow stress were compared with the experimental data.

2 Experimental

The chemical composition of 10Ni5CrMoV steel was as follows: C 0.09, Si 0.30, Mn 0.50, Cr 0.55, Ni 4.55, Mo 0.50, V 0.05, Al 0.021, Co 0.1, S 0.002, P 0.005, H 0.147, O 0.0024, N 0.0073, Fe balance. The initial grain size of the steel was 15 μm (about 9.5 grain number). The cylinder specimens with the size of 15 mm in height and 10 mm in diameter were electro-discharged from the initial material to carry out the hot compression experiments. The temperature at which the austenite phase was transformed into the ferrite and pearlite phase was 820 $^{\circ}\text{C}$.

The isothermal hot compression experiments were performed at a temperature range 800–1200 $^{\circ}\text{C}$, a strain rate range 0.01–10 s^{-1} , and a height reduction of 60% on a Gleeble-3500 simulator. Prior to the deformation, the specimens were heated up to the austenized temperature 1250 $^{\circ}\text{C}$ for 5 min to completely dissolve the precipitates. After the austenization, the specimens were rapidly cooled at 10 $^{\circ}\text{C s}^{-1}$ to the testing temperature and held for 5 s to ensure a uniform temperature distribution throughout the specimens. The deformation procedure is shown in Fig. 1. The deformation temperature was measured by a thermocouple that was spot-welded to the center of the specimen

Fig. 1 Experimental procedure of hot compression test for acquiring stress strain data and microstructure

surface. To reduce the die friction and ensure uniform deformation, two thin pieces of thallium were affixed to the flat dies. All of the compression tests were performed in an argon atmosphere. To preserve the microstructure at high temperatures, the compressed samples were water-quenched immediately after the compression test.

Following the hot compression experiments, the metallographic preparation of the specimens was carried out by cutting along their length axis, mechanically grinding using grit papers with different particle sizes, polishing and etching with a solution of 97% alcohol and 3% nitric acid for microstructural observation and with a solution of 5% picric acid, 5% BSNa, 1% hydrogen peroxide and distilled water for grain size statistic. Then, the optical micrographs were taken in the center zone of the specimens using an OLYMPUS PMG3 optical microscope. Three to five micrographs at each deformation condition were used to estimate the average grain size and the recrystallization volume fraction. The average grain size was measured and evaluated using the linear intercept method according to the ASTM E112-10 standard.

3 Results and Discussion

3.1 Flow Behavior

Typical stress–strain curves of the 10Ni5CrMoV steel during hot compression under temperatures from 800 to 1200 °C and strain rates from 0.01 to 10 s^{−1} are shown in Fig. 2. The stress was very sensitive to strain, strain rate, and temperature. The stress increased significantly with increasing strain during the initial stage and then reached a plateau or peak value at high strains under different deformation conditions. Moreover, the stress increased with increasing strain rate at a given temperature and decreased with increasing temperature at a given strain rate.

Figure 2 shows that the stress–strain curves exhibited an obvious softening tendency under some conditions. The stress increased with strain to peak values, after which the stress decreased gradually until it reached a steady state. This phenomenon may be caused by microstructural evolution, such as DRX or dynamic recovery (DV). At the strain rate of 10 s^{−1}, the stress increased with the strain. For stress that exhibited an obvious softening, a peak stress (σ_p) and its corresponding peak strain (ε_p) were easily found. These values are listed in Table 1. It was clearly seen that σ_p increased with decreasing temperature at a given strain rate and decreased with decreasing strain rate at a given temperature. From Table 1, we could see that the ε_p moved toward a higher strain level with decreasing deformation temperature.

To further quantify the softening mechanism, the relationships of softening extent ($\eta = \Delta\sigma/\sigma_p = (\sigma_p - \sigma_s)/\sigma_p \times 100\%$, where σ_s represents a stress at a steady stage) with temperature and strain rate are shown in Fig. 2f, which showed that $\Delta\sigma$ was sensitive to both temperature and strain rate. The softening extent increased with increasing temperature regardless of strain rate. It implied that the volume fraction of DRX increased with temperature, which resulted in a larger softening in flow stress. Moreover, the strain rate influenced softening extent greatly. This value decreased from 27 to 15% when strain rate increasing from 0.01 to 1 s^{−1}.

3.2 Constitutive Model

The deformation behavior of a material at high temperature can be described by an Arrhenius-type equation, which can give a good approximation among stress and strain, strain rate, and temperature [3]:

$$\dot{\varepsilon} = AF(\sigma) \exp\left(-\frac{Q}{RT}\right), \quad (1)$$

$$F(\sigma) = \begin{cases} A_1(\sigma)^{n_1} (\alpha\sigma < 0.8) \\ \exp(\beta_1\sigma) (\alpha\sigma > 1.2) \\ [\sinh(\alpha\sigma)]^n \text{ (for all } \sigma) \end{cases}, \quad (2)$$

where $\dot{\varepsilon}$ is the strain rate (s^{−1}), R is the universal gas constant (8.31 J mol^{−1} K^{−1}), T is the absolute temperature (K), Q is the activation energy for deformation (kJ mol^{−1}), and σ is the flow stress (MPa) for a given strain. A , A_1 , α , β_1 , n , and n_1 are material constants.

For all value of σ , the constitutive equation is written as:

$$\dot{\varepsilon} = A[\sinh(\alpha\sigma)]^n \exp\left(-\frac{Q}{RT}\right). \quad (3)$$

Taking the logarithm of both sides of Eq. (1) at the σ_p , the following formula can be obtained:

$$\ln \dot{\varepsilon} = n_1 \ln \sigma + \ln AA_1 - \frac{Q}{RT} (\alpha\sigma < 0.8), \quad (4)$$

and

$$\ln \dot{\varepsilon} = \beta_1 \sigma_p + \ln A - \frac{Q}{RT} (\alpha\sigma > 1.2). \quad (5)$$

According to Eqs. (4) and (5), the variations of flow stress σ_p with strain rate on a logarithm scale are shown in Fig. 3a and b to determine the values of n_1 and β_1 , respectively.

Figure 3a and b shows that the fitting lines were nearly parallel to each other; thus, the values of n_1 and β_1 were averaged to 7.63 and 0.10, respectively. According to the relation of $\alpha = \beta_1/n_1$, the value of α was taken as 0.013. It was assumed that the value of Q was independent of

Fig. 2 Stress–strain curves of steel at strain rates from 0.01 to 10 s^{-1} and temperatures of **a** 800 °C, **b** 900 °C, **c** 1000 °C, **d** 1100 °C, **e** 1200 °C, **f** softening extent

Table 1 Peak stresses and peak strains (σ_p/ϵ_p) at different conditions

Temperature (°C)	0.01 s^{-1}	0.1 s^{-1}	1 s^{-1}	10 s^{-1}
900	121.0/0.44	169.0/0.69	197.0/0.77	–
1000	76.4/0.22	116.0/0.43	152.0/0.71	–
1100	55.0/0.17	75.6/0.29	110.0/0.40	–
1200	40.0/0.12	55.0/0.20	75.0/0.33	–

temperature, so taking the logarithm of both sides of Eq. (3), it can be rewritten as:

$$\ln \dot{\epsilon} = n \ln[\sinh(\alpha\sigma)] + \ln A - Q/(RT). \quad (6)$$

According to Eq. (6), the relationships among $\ln \dot{\epsilon}$ and $\ln[\sinh(\alpha\sigma)]$, $\ln[\sinh(\alpha\sigma)]$ and $1/T$ are plotted in Figs. 4 and 5, respectively.

The value of n at different temperatures in Eq. (6) was determined by the slope of each fitted line in Fig. 6. The values of n were estimated to be 5.7, 4.3, 4.0, 4.2, and 5.1 at temperatures of 800, 900, 1000, 1100, and 1200 °C, respectively. The value of n varied slightly with temperature. The average value of n in Eq. (7) was 4.66. Similarly, the value of $Q/(nR)$ was determined by the slopes of the fitting lines in Fig. 5. The slopes of the fitting lines varied slightly from 9.3 to 10.3 as the strain rate ranged from 0.01 to 10 s^{-1} . Thus, the average value of Q was taken as 450 kJ mol^{-1} . The value of A was 2.06×10^{15} . Thus, the constitutive equation at peak strain can be written as:

$$\dot{\epsilon} = 2.06 \times 10^{15} [\sinh(0.01336\sigma)]^{4.66} \exp\left(-\frac{450,000}{RT}\right). \quad (7)$$

Fig. 3 Determining values of n_1 (a) and β_1 (b) from Eqs. (2) and (3)

Fig. 4 Determining value of n from Eq. (6)

Fig. 5 Determining value of Q from Eq. (6)

3.3 Predicting Flow Stress

The material constants in Eq. (3) were expressed as a function of strain, as depicted in Fig. 6. According to Eq. (3), the following formula can be obtained:

$$\left(\frac{Z}{A}\right)^{\frac{1}{n}} = \sinh(\alpha\sigma) = \frac{e^{\alpha\sigma} - e^{-\alpha\sigma}}{2}, \quad (8)$$

where Z represents the Zener–Hollomon parameter ($\dot{\epsilon} \exp(Q/(RT))$). By solving Eq. (8), the following formula can be obtained:

$$e^{\alpha\sigma} = \left(\frac{Z}{A}\right)^{\frac{1}{n}} + \left(\left(\frac{Z}{A}\right)^{\frac{2}{n}} + 1\right)^{\frac{1}{2}}. \quad (9)$$

Taking the logarithm of both sides of Eq. (9), the flow stress σ can be written as:

$$\sigma = \frac{1}{\alpha} \ln \left(\left(\frac{Z}{A}\right)^{\frac{1}{n}} + \left(\left(\frac{Z}{A}\right)^{\frac{2}{n}} + 1\right)^{\frac{1}{2}} \right). \quad (10)$$

By substituting the material constant functions (depicted in Fig. 6) into Eq. (10), the constitutive equation of the material for the entire strain range was obtained. Thus, the stress can be predicted, as shown in Fig. 7. The predicted stress matched well with the tested values, i.e., the predicted data increased with increasing strain rate and decreasing temperature.

To further evaluate the constitutive equation, standard statistical parameters, such as the correlation coefficient r and average absolute relative error (AARE), were employed, respectively, defined as [11]:

$$r = \frac{\sum_{i=1}^N [(E_i - \bar{E})(P_i - \bar{P})]}{\sqrt{\sum_{i=1}^N (E_i - \bar{E})^2 \sum_{i=1}^N (P_i - \bar{P})^2}}, \quad (11)$$

$$\text{AARE}(\%) = \frac{1}{N} \sum_{i=1}^N \left| \frac{E_i - P_i}{E_i} \right| \times 100, \quad (12)$$

where E represents the experimental data value and P is the corresponding predicted value, $\bar{E}$ and $\bar{P}$ are the average values of E and P , respectively, and N is the total number of data points. Figure 8 shows the correlation between the experimental and predicted values. A good correlation was obtained, i.e., the value of r was 0.977, while the largest AARE was 5.7%. This indicated that the constitutive equation was reliable and had good ability to predict the flow stress.

3.4 Processing Map

The processing map has been developed based on the dynamic material model, and it has been widely used to study the hot formability of metallic materials in the past years. The workpiece undergoing the hot deformation is considered to be a dissipater of power. An instantaneous power P_1 dissipated can be separated into two terms: G associated with the plastic deformation and J related to microstructural evolution. It is given by [24]:

$$P_1 = \sigma \dot{\epsilon} = G + J = \int_0^{\dot{\epsilon}} \sigma d\dot{\epsilon} + \int_0^{\sigma} \dot{\epsilon} d\sigma. \quad (13)$$

The strain rate sensitivity (m) of flow stress is the factor that partitions power between deformation heat and microstructural evolution [25]. The processing map consists of two parameters, namely efficiency of the power dissipation η and flow instability ζ . The η is defined as a function of temperature and strain rate given by:

$$\eta = \frac{J}{J_{\max}} = \frac{2m}{m+1}, \quad (14)$$

where m is the strain rate sensitivity, which is defined as:

Fig. 6 Variation of material constants in Eq. (3) with strain: **a** α ; **b** n ; **c** A ; **d** Q

Fig. 7 Comparison of predicted flow stress with experimental data at strain rates of **a** 0.01 s⁻¹, **b** 0.1 s⁻¹, **c** 1 s⁻¹, **d** 10 s⁻¹

Fig. 8 Correlation between experimental and predicted values

$$m = \left(\frac{\partial \ln \sigma}{\partial \ln \dot{\epsilon}} \right)_{T, \dot{\epsilon}} \quad (15)$$

The variation of η with temperature and strain rate represents a power dissipation map which is usually considered as an iso-efficiency contour map. A criterion for the onset of flow instability can be given by [26]:

$$\xi = \frac{\partial \ln \left(\frac{m}{m+1} \right)}{\partial \ln(\dot{\epsilon})} + m < 0. \quad (16)$$

The negative value represents a regime of instability. A superimposition of the instability map on the power dissipation map gives a processing map. By the processing map, safe regions with efficiency contours converging toward a peak efficiency and unsafe regions with ξ less than zero can be identified. For the safe regions, the intrinsic workability of the material is good and such as DRX or superplasticity may occur. While for the unsafe regions, flow instability may occur, i.e., adiabatic shear bands or flow localization. Therefore, an appropriate temperature–strain rate processing window for the hot working can be found. The processing maps for the material at different strains are shown in Fig. 9.

In Fig. 9, the contour numbers represented the power dissipation efficiency and the shaded domains indicated the unsafe regions of flow instability (negative ξ). At a strain of 0.2 (Fig. 9a), there was a small shaded domain at low temperature and high strain rate, while the maximum power dissipation efficiency reached 0.28. This indicated that flow instability may not occur under these conditions, which may result from the lower plastic deformation that could not produce defects such as adiabatic shear bands or localized plastic flow. As the deformation continued (Fig. 9b), the strain increased to 0.4, the shaded region began to expand, and flow instability may occur. With further strain to 0.8, the maximum η reached 0.46 and was located in the domain at higher temperatures (1100–1200 °C) and lower strain rates (near 0.001 s^{−1}). Under these conditions, the microstructures had obviously

changed. The typical microstructural evolution was DRX. This phenomenon was validated by the following factors. The original grain size was refined, the grain boundaries became curved, and many new equiaxed grains appeared along original boundaries, as shown in Fig. 10.

3.5 Microstructural Evolution

3.5.1 Dynamic Recrystallization

Figure 11 shows the microstructural evolution of the material at different conditions. The effects of temperature and strain rate were significant. The original grains were elongated perpendicular to the compression direction at 800 °C (Fig. 11a–c). With the increase in temperature, elongated grains were gradually replaced by equiaxed grains. Typical DRX also took place. The new recrystallized grains preferably formed around the original grain boundaries. This was mainly because the zones around grain boundaries had many defects with high deformation energy; thus, nucleation occurred easily and more recrystallized grains appeared in these zones.

The recrystallized grain size and volume fraction of recrystallization (f_{drx}) were sensitive to the strain rate and temperature. These values increased with increasing temperature and decreasing strain rate; for example, the recrystallized grain size and the volume fraction at 1200 °C decreased from 100 to 35 μm, and from nearly 100–80%, respectively, when the strain rate increased from 0.01 to 1 s^{−1}; at a strain rate of 0.01 s^{−1}, these values increased from 5 to 100 μm, and from 5 to 100%, respectively, when the temperature increased from 800 to 1200 °C.

The variation of recrystallized grain size with temperature and strain rate can be explained by the following reasons. It is known that the size is determined by the migration mobility of the grain boundary ($v = MP$, where v is migration rate, M is migration mobility, and P is the driving force), with the driving force P provided by the difference in dislocation density and growth time. Under the same strain rate, the migration mobility increases with temperature, while the driving force decreases. However, the reduction in the driving force is smaller than the increase in migration mobility; therefore, the growth velocity increases with temperature. At a given temperature, although the driving force of a recrystallized grain increases with increasing strain rate, the growth time is largely reduced. Therefore, the recrystallized grain size decreased with increasing strain rate.

3.5.2 Critical Condition for Onset of Dynamic Recrystallization

The stress–strain curves of the material exhibited an obvious softening trend, which mainly resulted from DRX.

Fig. 9 Processing maps of steel deformed at different strains of **a** 0.2, **b** 0.4, **c** 0.8

Fig. 10 Typical microstructures of steel showing DRX deformed at strain rate of 0.1 s^{-1} and higher temperatures of **a** 950 °C, **b** 1000 °C

The onset of DRX can therefore be determined by these curves. However, DRX usually initiates at a critical condition (strain or stress) before the peak value. The critical condition depends on the basic chemical composition and thermo-mechanical parameters of the material. Many researchers have proposed various mathematics models to predict the critical condition. Based on the static recrystallization nucleation mechanism, Barnett et al. [27] developed a model to identify a particular strain as the critical condition for the initiation of DRX, while Gottstein et al. [28] identified the critical strain based on a dislocation work-hardening model. Poliak and Jonas [26] believed that the critical strain for the initiation of DRX can be directly identified from the inflections in curves of $(-\partial\theta/\partial\sigma) - \sigma$, where θ represents the work-hardening rate ($\theta = \partial\sigma/\partial\varepsilon$, where σ is stress and ε is strain). Najafizadeh and Jonas [19] subsequently simplified the model of Poliak and Jonas and proposed a method to fit the strain-hardening data to a third-order polynomial to determine the critical condition. Recently, Chen et al. [18] concluded that considering the work-hardening rate as a function of stress was

the best method to determine the critical condition for the onset of DRX. In this work, we adopted this method to determine the critical condition. The curves of $\theta - \sigma$ were fitted to a third-order polynomial expressed as:

$$\theta = B_0 + B_1\sigma + B_2\sigma^2 + B_3\sigma^3, \quad (17)$$

where B_0 to B_3 are material constants. To determine the inflection point in the curves, the second derivative of Eq. (17) with respect to σ was taken, expressed as:

$$\frac{d^2\theta}{d\sigma^2} = 2B_2 + 6B_3\sigma. \quad (18)$$

Setting Eq. (18) equal to zero, the value corresponds to the critical stress for the initiation of DRX. Thus,

$$2B_2 + 6B_3\sigma = 0 \text{ and } \sigma_c = -\frac{B_2}{3B_3}, \quad (19)$$

where σ_c is the critical stress. After determining σ_c , the corresponding critical strain (ε_c) can be identified from the stress-strain curves. Figure 12a–c shows the third-order fitting curves of $\theta - \sigma$. The fitted curves matched the

Fig. 11 Microstructural evolution of steel at temperatures of **a–c** 800 °C, **d–f** 1000 °C, **g–i** 1200 °C and strain rates of **a, d, g** 0.01 s^{−1}, **b, e, h** 0.1 s^{−1}, **c, f, i** 1 s^{−1}

experimental data well. According to Eq. (19), σ_c was obtained from the derivative of the fitted third-order polynomials. Figure 12d–f shows the relations of $(-\partial\theta/\partial\sigma) - \sigma$, which were used to determine the critical stress. The corresponding ε_c was determined from the stress–strain curves. The values of critical stress σ_c and critical strain ε_c are listed in Table 2.

3.5.3 Dynamic Recrystallization Kinetics

DRX is a typical thermally activated process that takes place via nucleation and subsequent growth. It is generally accepted that only when strain reaches the critical value can nucleation occur. With increasing strain, the recrystallized nuclei begin to grow by grain boundary migration. The nucleation and growth process can be considered to represent the kinetics of DRX. The basic characteristic of DRX kinetics displays a sigmoidal form with respect to time. This means that an initial incubation period is

followed by a rate of change that is initially slow, then accelerates, and, finally, becomes slow again as the process is completed. DRX kinetics at a constant strain rate can be expressed as [18]:

$$f_{\text{drx}} = 1 - \exp \left[-\beta_d \left(\frac{\varepsilon - \varepsilon_c}{\varepsilon_{0.5}} \right)^{k_3} \right] \quad (\varepsilon > \varepsilon_c), \quad (20)$$

where f_{drx} is the DRX volume fraction, β_d and k_3 are the material constants, ε and ε_c are the strain and critical strain, respectively. $\varepsilon_{0.5}$ is the strain to 50% DRX volume fraction, which is written as:

$$\varepsilon_{0.5} = k_1 \dot{\varepsilon}^{k_2} \exp \left(\frac{Q_{\text{drx}}}{RT} \right), \quad (21)$$

where k_1 and k_2 are material constants and Q_{drx} is the activation energy for DRX (kJ mol^{−1}). The value of f_{drx} can be determined by either quantitative metallographic measurement or from a flow-curve analysis. Flow-curve analysis is a simpler and quicker method, because

Fig. 12 Curves of $\theta - \sigma$ (a–c) and $(-\partial\theta/\partial\sigma) - \sigma$ (d–f) at strain rates of a, d 0.01 s^{-1} , b, e 0.1 s^{-1} , c, f 1 s^{-1}

Table 2 Values of critical strain and stress (σ_c/ε_c) under different deformation conditions

Temperature ($^{\circ}\text{C}$)	0.01 s^{-1}	0.1 s^{-1}	1 s^{-1}	10 s^{-1}
900	80.0/0.07	—	—	—
1000	60.1/0.10	93.5/0.14	—	—
1100	47.8/0.09	63.0/0.11	94.6/0.20	—
1200	34.8/0.06	40.0/0.06	66.0/0.16	—

quantitative metallographic measurement needs a large quantity of samples to be examined and it is difficult to precisely ascertain the new grains [13]. The method for determining f_{drx} by analyzing the flow stress curves is illustrated in Fig. 13. In this study, we adopted the following expression to determine f_{drx} :

$$f_{\text{drx}} = \frac{\Delta\sigma_s}{\sigma_{\text{sat}} - \sigma_{\text{ss}}}, \quad (22)$$

where $\Delta\sigma_s$ is considered as the net softening effect directly attributable to DRX. This is expressed as:

$$\Delta\sigma_s = \sigma_{\text{sat}} - \sigma, \quad (23)$$

where σ_{sat} denotes the flow stress and its softening effect that only results from dynamic recovery (without DRX softening). σ_{ss} is the steady-state stress under the DRX conditions.

According to Eq. (22), the value of f_{drx} can be determined from the stress-strain curves. The method to calculate the values of σ_{sat} and σ_{ss} can be found in Ref. [29].

Fig. 13 Method to determine volume fraction of DRX through net softening effect

The value of f_{drx} determined by Eq. (22) is shown in Fig. 14. f_{drx} showed a sigmoidal form with respect to time and was very sensitive to temperature and strain rate. With a decrease in temperature or increase in strain rate, the value of f_{drx} became smaller, while the incubation period became longer.

To determine the material constants in Eq. (20), the logarithms of both sides of Eq. (21) were taken:

$$\ln \varepsilon_{0.5} = \ln k_1 + k_2 \dot{\varepsilon} + \frac{Q_{\text{drx}}}{RT}. \quad (24)$$

Thus, by linear regression of Eq. (24), the values of k_2 and Q_{drx} can be taken as 0.2 and $65,373.38 \text{ J mol}^{-1}$, respectively, as displayed in Fig. 15a and b. Substituting these values into Eq. (24) gave a value of k_1 of 0.002.

Fig. 14 DRX kinetics at strain rates of **a** 0.01 s⁻¹, **b** 0.1 s⁻¹, **c** 1 s⁻¹

Similarly, taking the logarithmic form of Eq. (20), we can obtain:

$$\ln(-\ln(1-f_{\text{drx}})) = \ln \beta_d + k_3 \ln\left(\frac{\varepsilon - \varepsilon_c}{\varepsilon_{0.5}}\right). \quad (25)$$

The relationship between $\ln(-\ln(1-f_{\text{drx}}))$ and $\ln[(\varepsilon - \varepsilon_c)/\varepsilon_{0.5}]$ is depicted in Fig. 15c. The values of k_3 and β_d were obtained by linear regression as 2.43 and 1.07, respectively. Finally, the kinetics equation of DRX can be expressed as:

$$f_{\text{drx}} = 1 - \exp\left[-1.07\left(\frac{\varepsilon - \varepsilon_c}{\varepsilon_{0.5}}\right)^{2.43}\right] (\varepsilon > \varepsilon_c). \quad (26)$$

To validate the DRX kinetics, the predicted f_{drx} was compared with the experimental data, as shown in Fig. 16. The trend predicted by Eq. (26) exhibited obvious sigmoidal characteristics, which was similar to that of the experimental data. The predicted values increased with increasing temperature and decreasing strain. During the initial stages, the predicted data increased slowly with strain, but grew rapidly until a steady stage, when the DRX was nearly complete.

To further validate the kinetics model, the flow stress can be modified as [29]:

$$\sigma = (1 - f_{\text{drx}})\sigma_{\text{sat}} + f_{\text{drx}}\sigma_{\text{ss}}. \quad (27)$$

Fig. 15 Determining values of **a** k_1 , **b** k_2 , **c** β_d and k_3

Fig. 16 Comparison of predicted value with experimental data at strain rates of **a** 0.01 s⁻¹, **b** 0.1 s⁻¹, **c** 1 s⁻¹

Fig. 17 Comparison of predicted flow stress with experimental data at strain rates of **a** 0.01 s^{-1} , **b** 0.1 s^{-1} , **c** 1 s^{-1}

The flow stress predicted by Eq. (27) is shown in Fig. 17. The predicted values agreed well with the experimental data.

4 Conclusions

- (1) The deformation behavior and DRX of 10Ni5Cr-MoV steel were experimentally investigated. A reliable Arrhenius-type constitutive model was established. To validate the model, the predicted flow stress was compared with the experimental values: the correlation coefficient and the largest relative error were 0.977 and 5.7%, respectively.
- (2) Processing maps at strains of 0.2, 0.4, and 0.8 were constructed, which were used to identify the processing conditions where flow instability occurred.
- (3) The critical strain for the onset of DRX was identified. A DRX kinetics model was developed that had good ability to predict the recrystallization volume fraction.

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