Hot Deformation Behavior of an Ultra-High-Strength Fe–Ni–Co-Based Maraging Steel

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Abstract

Hot processing behavior of an ultra-high-strength Fe–Ni–Co-based maraging steel was studied in temperature range of 900–1200 °C and strain rate range of 0.001–10 s^{−1}. Deformation processing parameters and optimum hot working window were characterized via flow stress analysis, constitutive equation construction, hot processing map calculation and microstructure evolution, respectively. Critical strain value for dynamic recrystallization was determined through theoretical mathematical differential method: the inflection point of θ – σ and $-\partial\theta/\partial\sigma$ – σ curves. It was found that the flow stress increased with the decrease in deformation temperature and increase in the strain rate. The power dissipation maps in the strain range of 0.1–0.6 were entirely similar with the tendency of contour lines which implied that strain had no strong effect on the dissipation maps. Nevertheless, the instability maps showed obvious strain sensitivity with increasing strain, which was ascribed to the flow localization and instability. The optimized hot processing window of the experimental steel was obtained as 1100–1200 °C/0.001–1 s^{−1} and 1000–1100 °C/0.001–0.1 s^{−1}, with the efficiency range of 20–40%. Owing to high Mo content in the experimental steel, high dynamic activation energy, $Q = 439.311$ kJ mol^{−1}, was achieved, indicating that dynamic recrystallization was difficult to occur in the hot deformation process, which was proved via microstructure analysis under different hot deformation conditions.

Keywords Maraging steel · Hot working · Dynamic recrystallization · Constitutive equation · Hot processing map

1 Introduction

Maraging steel (martensite aging steel) is regarded as an important material candidate for critical applications such as rocket engine cases, submarine hulls, cryogenic missiles, atomic energy equipment and other fields due to its super high strength and toughness [1–4]. The high strength of

maraging steel is mainly achieved by the precipitation of intermetallic compounds during an appropriate aging treatment, and the sufficient toughness and ductility could be attributed to the soft martensite matrix composed of a very low carbon content (generally less than 0.003 wt%) [3, 5–8].

In the past, researches on maraging steels were mainly focused on the strengthening mechanism, precipitation behavior and heat treatment [3, 5, 9, 10]. Less work has been done on the hot deformation behavior and the microstructure evolution of this class of materials [11, 12]. The processing map (PM) technology based on the dynamic materials model (DDM) is a powerful approach to analyze the deformation mechanisms and optimize the processing parameters of materials due to its convenience and accuracy [13]. Therefore, in order to improve the synthetic properties of maraging steels, it is of great theoretical and practical significance to conduct a systematic study on the effect of hot working parameters on the hot deformation behavior and the microstructure evolution of maraging steels.

PM has been used in a wide range of materials including aluminum alloy, titanium, magnesium, Ni-based alloy and

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steels to optimize their processing parameters and investigate their microstructure evolutions [14–18]. Furthermore, PM has been used to identify the instability zones and the workability area combined with the microstructure analysis [19, 20]. A necessary tool for quantitative analysis of the relationship among the hot deformation parameters in the hot working procedure, such as stress, strain, strain rate and temperature, is the constitutive equation [21, 22]. The constitutive equation can be used not only to describe the flow stress response, but also to predict the deformation behavior under some uncertain conditions [23, 24].

In the present work, the hot working behavior of a Fe–Ni–Co-based maraging steel (MAS) and the effect of hot deformation parameters on flow stress curve and microstructure evolution were investigated. Flow stress curves, work-hardening curves, constitutive equations, PM and microstructure analysis were used to examine the hot deformation characteristics and to determine the optimum hot processing parameters for the present steel.

2 Experimental

The experimental steel was initially melted in a 200 kg vacuum induction melting furnace and then re-melted in a vacuum arc re-melting furnace. After homogenization at 1250 °C for 24 h, the re-melted ingot was forged into rods with 22 mm in diameter followed by air cooling, and then the rods were machined into hot compression specimens with 8 mm in diameter and 12 mm in thickness. The chemical composition (wt%) of MAS is as follows: 0.003C, 0.0018P, 0.002S, 0.0014O, 0.0026 N, 17.7Ni, 15.0Co, 6.6Mo, 1.2Ti, 0.07Al and Fe in balance, and the microstructure of MAS at the initial state is shown in Fig. 1. It

Fig. 1 Microstructure of the experimental MAS at the initial state

can be found that the steel matrix is a typical martensite characteristic with grain size of about 70 μm.

Hot compression tests were conducted on a Gleeble 3800 thermal simulation machine at temperatures of 900, 1000, 1050, 1100 and 1200 °C, and strain rates of 0.001, 0.01, 0.1, 1 and 10 s⁻¹, respectively. Graphite and MoS₂ pastes were used to minimize the friction between the specimen and the platen. A chromel–alumel-type thermocouple was embedded in the middle of the specimen to monitor the temperature during the whole hot compression procedure.

Each specimen was first inductively heated to 1200 °C and held for 5 min, then cooled down to the desired deformation temperature at a cooling rate of 10 °C s⁻¹. To obtain a stable and uniform temperature prior to deformation, isothermal holding for 1 min was performed at each deformation temperature. All the specimens were compressed to a true strain of 0.9 and then water-quenched to room temperature. Figure 2 shows the schematic diagram of whole experimental procedures.

The compressed specimens were sectioned along the loading direction and then their surfaces were ground and electro-polished using 10% perchloric acid solution to observe the microstructure evolution using a scanning electron microscope (SEM, TESCAN MAIA3 120 kV) equipped with electron back-scattered diffraction (EBSD, Oxford NordlysMax3). In order to obtain the detailed evolution information of the martensite microstructure, EBSD was performed with scan area 130 μm × 100 μm and step size 0.3 μm. All the results based on EBSD analysis were obtained by HKL Channel 5.

Fig. 2 Schematic diagram of hot compression test

3 Results and Discussion

3.1 Flow Stress Behavior

The true stress–strain curves of the experimental steel in the temperature range of 900–1200 °C and strain rate of 0.001–10 s^{−1} are shown in Fig. 3. It can be observed from Fig. 3 that the flow stress is strongly dependent on the deformation temperature and strain rate. The flow stress increases with increasing strain rate at a given deformation temperature and decreases with increasing deformation temperature at a given strain rate. Figure 3a shows that the flow stress increases rapidly with increasing strain rate when the deformation temperature is 900 °C. However, when the strain rate is 0.01 s^{−1} (Fig. 3f), the flow stress decreases with increasing deformation temperature which is because the high deformation temperature can increase the vacancy diffusion rate, and promote cross-slipping of screw dislocations and climbing of edge dislocations [25]. Additionally, all the flow stress curves increase rapidly at the initial stage of hot deformation, which can be attributed to the work hardening. But some of the curves show a distinct peak such as at 1100 °C and 0.001 s^{−1}, indicating the onset of dynamic recrystallization (DRX).

When the deformation temperatures and strain rates are 1050 °C, 1100 °C, 1200 °C and 0.01 s^{−1} and 0.001 s^{−1}, respectively, the flow curves show a single peak followed by a decrease in stress and finally reaching a plateau, which implies the occurrence of DRX, as shown in Fig. 3c–e. But when the strain rates are 1 s^{−1} and 10 s^{−1}, the flow stress curves show a continuous rise character without peak stress, indicating that the work-hardening mechanism is dominant and softening mechanisms such as DRV and DRX have slight effect on the deformation [14, 26]. However, when the deformation temperatures are 900 °C and 1000 °C, the flow stress curves show continuous rise features without a softening, which can be explained that at lower temperature or higher strain rate, there is insufficient thermodynamic condition or time for both nucleation and growth of DRX grains and dislocation annihilation [26, 27].

3.2 Constitutive Analysis

3.2.1 Work-Hardening Curves

Generally, the critical strain (ϵ_c) indicates the onset of DRX. Compared to the microstructure observation which is one of methods used to identify ϵ_c , flow stress curve analysis is considered to be the most convenient and accurate way to

Fig. 3 True stress–strain curves of the experimental steel at different strain rates and temperatures of **a** 900 °C, **b** 1000 °C, **c** 1050 °C, **d** 1100 °C, **e** 1200 °C, **f** flow stress curves at different temperatures with strain rate of 0.01 s^{−1}

calculate ε_c . The flow stress curve method was first proposed by Mecking and Kocks [28], and then further developed by McQueen and Ryan [29, 30]. This method provides a more effective way to focus on details of hot working curves. In the light of the differential theories, the shape of the flow stress curve and the associated work-hardening rate curve, $\theta = \partial\sigma/\partial\varepsilon$, can be affected and some relative inflection points can be identified as the critical point for the onset of DRX.

Mathematical flow stress curve analysis method has been used to define the critical strain for the onset of DRX. The work-hardening rate, $\theta = \partial\sigma/\partial\varepsilon$, is obtained from the flow stress curve and plotted as a function of stress, as shown in Fig. 4. According to the work by Zahiri et al. [31], critical stress, σ_c (the onset of the deviation), peak stress, σ_p (the intersection point of curve and horizontal coordinate) and saturation stress, σ^*_s (the intersection of the extrapolation of the starting point to the horizontal axis, dynamic recovery (DRV) as the unique softening mechanism), can be determined from the work-hardening curves.

Work-hardening curves of the experimental steel showed similar feature as other researchers [17, 18]. All the work-hardening curves decreased sharply with increasing stress until an inflection point at which the slope reduction slowed down, indicating that the DRX began to occur. By extrapolating the inflection point to the horizontal coordinate, σ^*_s can be obtained as shown in Fig. 4. DRV is the main factor leading to the rapid decline at the initial stage [27].

Although the first-order differential method can clearly define the position of the inflection point, it is difficult to obtain the exact value accurately. Poliak et al. [32] proposed a mathematical method which uses the second-order differential to determine the accurate inflection point. In this method, the second-order differential calculation of flow stress as a function of stress ($-\partial(\frac{\partial\theta}{\partial\sigma})/\partial\sigma$ vs. σ) shows a single minimum

Fig. 4 Variation of work-hardening rate with flow stress under different deformation conditions

point at the inflection on the work-hardening curve. As shown in Fig. 5, the accurate value of the inflection point can be recognized clearly and efficiently. Thus, the relevant critical stress and critical strain can be obtained from the original flow stress curve. In Fig. 5, the strain hardening rate Γ_c ($\Gamma_c = -\ln\sigma/\partial\varepsilon$), which has been described and explained in detail by Poliak [32], could be obtained by calculation. In the current work, the value of Γ_c was 0.585, which is different from that of $\Gamma_c = 0.27$ reported by McQueen and Ryan [29, 30] and is quite similar to that of $\Gamma_c = 0.6$ reported by Poliak et al. [32]. This difference or similarity of the Γ_c value is probably attributed to the chemical composition and deformation mode of the material used.

3.2.2 Establishment of Constitutive Equation

Constitutive equation plays a significant role in predicting and analyzing the relationship among the deformation temperature, the strain rate and the flow stress in the plastic deformation process of metallic materials. Among all the equations, Arrhenius-type equation proposed by Sellars et al. [33] has been widely used for engineering materials. Additionally, the Z parameter method has also been widely used to analyze the relationship among the hot processing parameters [34]. These equations are listed as follows:

$$\dot{\varepsilon} = A_1 \sigma^{n_1} \exp\left(-\frac{Q}{RT}\right). \quad (1)$$

$$\dot{\varepsilon} = A_2 \exp(\beta\sigma) \exp\left(-\frac{Q}{RT}\right). \quad (2)$$

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

Fig. 5 Variation of $-\partial\theta/\partial\sigma$ as a function of flow stress under different conditions

$$Z = \dot{\epsilon} \exp\left(\frac{Q}{RT}\right). \quad (4)$$

where $\dot{\epsilon}$ is the strain rate, σ is the flow stress, Q is the apparent activation energy of deformation, R is the universal gas constant ($8.314 \text{ J mol}^{-1} \text{ K}^{-1}$), T is the absolute temperature, A , A_1 , A_2 , n_1 , n , β and α are material constants.

Generally, the power law description of stress [Eq. (1)] is preferred for relatively low stress, while the exponential law [Eq. (2)] is suitable for high stress conditions. While, the hyperbolic sine law [Eq. (3)] is applicable in a wide range of deformation temperature and strain rate [35]. So, all the material constants in the equations mentioned above need to be identified to further analyze the constitutive equations of the experimental steel. So far, material constant α in the constitutive equation which has no physical meaning could be obtained mainly in two methods, which has been reported in many references [30, 41–44]. The first method was developed by Uvira and Jonas in the 1960s. In this method, the relation between n and α was depicted directly at a constant temperature according to the hyperbolic sine equation, and the standard deviation of n vs α was plotted to get the minimum n . Finally, α was obtained according to the minimum n . The second method, used in this study, has been widely used in many metallic materials, like stainless steel [45], aluminum alloy [46], copper

alloy [47], etc. In this method, material constants, β and n_1 , usually were calculated by linear regression according to the exponential and power law. Then, the value of α could be obtained by the equation, $\alpha \approx \beta/n_1$. Even though these two methods are different in calculation processes, they do have some similarities in calculation principles which aim to obtain the value of α . According to the first method, the minimum standard deviation of n was calculated to be 0.445 which means that n varies greatly with the temperature. It was thought that the difference may be caused by the different selection method of stress values while calculating n values. Consequently, the latter method was selected to calculate the value of α .

Taking natural logarithm of Eqs. (1) and (2) yields:

$$\ln \sigma = \frac{\ln \dot{\epsilon}}{n_1} - \frac{\ln A_1}{n_1} - \frac{Q}{n_1 RT}. \quad (5)$$

$$\sigma = \frac{\ln \dot{\epsilon}}{\beta} - \frac{\ln A_2}{\beta} + \frac{Q}{\beta RT}. \quad (6)$$

Characteristic stresses such as peak, steady or the stress corresponding to a specific strain can be used for the calculation of material constants. In this study, the peak stress values were used to calculate A , n_1 , β , α and Q , as shown in Fig. 6. n_1 and β could be obtained from the slope of the linear regression lines by plotting $\ln \sigma - \ln \dot{\epsilon}$ and $\sigma - \ln \dot{\epsilon}$, as

Fig. 6 Relationship among flow stress, strain rate and deformation temperature: **a** $\ln \sigma - \ln \dot{\epsilon}$, **b** $\sigma - \ln \dot{\epsilon}$, **c** $\ln[\sinh(\alpha\sigma)] - \ln \dot{\epsilon}$, **d** $\ln[\sinh(\alpha\sigma)] - 1000/T$, **e** $\ln[\sinh(\alpha\sigma)] - \ln Z$

shown in Fig. 6a, b respectively. The average values of n_1 and β were calculated to be 6.62 and 0.0456 MPa^{-1} , then $\alpha \approx \beta/n_1 = 0.0069 \text{ MPa}^{-1}$.

Taking natural logarithm of Eq. (3) yields:

$$\ln \dot{\epsilon} = \ln A + n \ln[\sin(\alpha\sigma)] - \frac{Q}{RT}. \quad (7)$$

Re-arranging Eq. (7) yields:

$$\ln[\sin(\alpha\sigma)] = \frac{\ln \dot{\epsilon}}{n} + \frac{Q}{nRT} - \frac{\ln A}{n}. \quad (8)$$

By plotting $\ln[\sinh(\alpha\sigma)] - \ln \dot{\epsilon}$ at constant temperature, the material constant n can be calculated by the slope of linear regression lines, as shown in Fig. 6c. The average value of n is 4.808 from the slope of four linear regression lines at different temperatures. Therefore, in the same way, the apparent activation energy of $439.311 \text{ kJ mol}^{-1}$ could be calculated from the average slope of the $\ln[\sinh(\alpha\sigma)] - 1/T$, as shown in Fig. 6d. The value of A can be obtained from the slope of the $\ln[\sinh(\alpha\sigma)] - \ln Z$, as shown in Fig. 6e, and A is about 0.9916×10^{18} in the current work. By filling in all the obtained material constants into Eqs. (3) and (4), the hyperbolic sine equation embracing the Z parameter for the experimental steel is developed as follows:

$$Z = \dot{\epsilon} \exp\left(\frac{439311}{RT}\right) = 0.9916 \times 10^{18} [\sinh(0.0069 \times \sigma)]^{4.808}. \quad (9)$$

According to present results, the activation energy of the experimental steel is $439.311 \text{ kJ mol}^{-1}$ which is much greater than that of the self-diffusion in γ -iron ($Q = 280 \text{ kJ mol}^{-1}$, indicating that the DRV and DRX are the dominant mechanisms instead of diffusion during the hot deformation process [36]. Compared with the ultra-high-strength steel, such as AF1410, the Q for the experimental steel is higher than that of AF1410 ($366.89 \text{ kJ mol}^{-1}$) [37], implying that the deformation resistance for the experimental steel is higher than AF1410, which is mainly due to high content of Mo. It was reported that steels containing Mo usually have higher dynamic recrystallization activation energy for the high self-diffusion coefficient of Mo [38, 39].

3.3 Processing Map

3.3.1 Power Dissipation Map

PM plays a very important role not only in investigating and predicting the dynamic behavior and hot workability of metals under hot deformation, but also in optimizing processing parameters and guiding industrial production. Based on the theory of DMM, PM was firstly proposed by Prasad et al. [13], which has been described in detail in the previous work. According to DMM, the whole hot deformation

procedure could be considered as a thermodynamically closed system and the total power of the system (P) could be divided into two parts: one is the energy consumed by plastic deformation of the material (G), and the other is the power consumed by microstructure evolution during hot deformation (J). The relation among the three parameters is written as follows:

$$P = G + J. \quad (10)$$

The ratio of these two energies is determined by the strain rate sensitivity index (m) of the material under a certain stress level, which is described as:

$$m = \frac{\partial J}{\partial G} = \frac{\dot{\epsilon} \partial \sigma}{\sigma \partial \dot{\epsilon}} = \frac{\partial(\ln \sigma)}{\partial(\ln \dot{\epsilon})}. \quad (11)$$

The parameter η is termed as a non-dimensional parameter, indicating the efficiency of power dissipation during the microstructure evolution. Therefore, η is thought to be a predictive index of microstructure change and the hot workability of material under hot deformation. It is defined as:

$$\eta = \frac{2m}{m+1}. \quad (12)$$

where m is the strain rate sensitive index.

Power dissipation map is composed by deformation temperature, strain rate and efficiency variation of power dissipation. Different domains in the map are related to various microstructure evolutions, such as DRV, DRX and super-plasticity. Generally, a higher value of η represents the more efficient processing parameters. The power dissipation maps of the experimental steel in the temperature range of $900\text{--}1200^\circ\text{C}$ and strain rate range of $0.001\text{--}10 \text{ s}^{-1}$ at true strains of 0.1, 0.2, 0.3, 0.4, 0.5 and 0.6 are shown in Fig. 7, respectively. The values of the contour numbers used in microstructure evolution during the hot working process represent the efficiency of power. It can be found from Fig. 7 that η changes slightly with increasing true strain, and the maximum value of η is between 0.36 and 0.40. It is highlighted that the density of contour lines becomes denser in the temperature range of $1000\text{--}1100^\circ\text{C}$ and strain rate of $0.01\text{--}0.1 \text{ s}^{-1}$, indicating that both temperature and strain rate have great effect on η . Under relative high power dissipation, more energy might be used to promote the evolution of microstructure, for example, DRV and DRX.

3.3.2 Processing Map

Flow instability, for instance, crack, adiabatic shear band, flow localization and second-phase precipitation, might take place in the hot processing procedure, which could have detrimental effect on the deformation process. In order to avoid or predict the flow instability, the flow instability criterion,

Fig. 7 Power dissipation maps at true strains of **a** 0.1, **b** 0.2, **c** 0.3, **d** 0.4, **e** 0.5, **f** 0.6

ξ , which was proposed by Ziegler [40], is used in the present study, and it can be depicted as:

$$\xi = \frac{\partial \ln[m/(m+1)]}{\partial \ln \dot{\epsilon}} + m \leq 0. \quad (13)$$

A negative value of ξ indicates an unstable condition for plastic deformation. The stability parameter is calculated as a function of deformation temperature and strain rate at a given strain, and can obtain a flow instability map.

Figure 8 shows PM of the experimental steel at true strains of 0.1, 0.2, 0.3, 0.4, 0.5 and 0.6, respectively. The shaded regions in the PM represent the instability regions and the contour number represents the efficiency of power dissipation. It can be seen from Fig. 8 that when the true strains are 0.1 and 0.2, there are two similarly unstable regions, 900–1000 °C at a relatively low strain rate of 0.001–1 s^{−1} and 1050–1100 °C at a high strain rate of 1–10 s^{−1}. When the true strain increases to 0.3 and 0.4, the instability region changes to a large region, 900–1000 °C at strain rate of 0.01–10 s^{−1} and 1050–1100 °C at 0.1–10 s^{−1} (as shown in Fig. 8c, d), respectively. As the true strain increases to 0.5 and 0.6, the instability maps tend to be similar, which can be divided into three consecutive stages: low temperature stage (900–1000 °C at strain rate of 0.01–10 s^{−1}), middle temperature stage (1000–1100 °C at strain rate of 0.1–10 s^{−1})

and high temperature stage (1100–1200 °C at strain rate of 1–10 s^{−1}). It can be concluded that as the deformation temperature increases, the strain rate corresponding to the unstable region becomes higher and the instability maps trend to be the same when the strain continues to increase. Therefore, in this work, the optimum hot working condition of the experimental steel for all the true strains occurs mainly in two regions: the strain rate range of 0.001–1 s^{−1} and temperature range of 1100–1200 °C and the strain rate range of 0.001–0.1 s^{−1} and temperature range of 1000–1100 °C. If 1200 °C was chosen as the deformation temperature, there is a wider selection zone of strain rate in the range of 0.001–10 s^{−1} and at the same time the emergence of unstable area can be avoided.

3.3.3 Microstructure Evolution

As mentioned above, when 1200 °C was selected as deformation temperature, any strain rate could be chosen as an alternative process parameter. However, there are significant differences about energy dissipation factors in different locations of the safe zone. For example, the dissipation factors at 1200 °C at true strain of 0.4 and 0.6 vary from 0.2520 to 0.4035 and from 0.2515 to 0.3824, respectively. The energy dissipation factors represent energy dissipation of microstructure evolution during hot working process, while the

Fig. 8 Hot processing maps at true strains of **a** 0.1, **b** 0.2, **c** 0.3, **d** 0.4, **e** 0.5, **f** 0.6

Fig. 9 Distribution maps of inverse pole figure-Z at 1200 °C under different strain rates at true strain of 0.6: **a** 0.001 s⁻¹, **b** 0.01 s⁻¹, **c** 0.1 s⁻¹, **d** 1 s⁻¹, **e** 10 s⁻¹

difference of energy dissipation factors in different safety zones represents the difference of microstructure evolution characteristics. Therefore, study on the microstructure evolution in the safety zone of different power dissipation factors is important to understand the microscopic mechanism during thermal deformation process. Figure 9 shows the microstructures of the specimens deformed at 1200 °C with true strain of 0.6 of different strain rate.

It can be seen from Fig. 9 that the width of martensite laths of specimen at lower strain rate is larger than that of

under higher strain rate. Meanwhile, the martensite laths in one grain show their own merit-based orientation, which can be attributed to the large strain capacity. With increasing strain rate, small amount of DRX grains are visible. To further analyze the DRX feature and microstructure evolution, the distribution maps of different microstructures are depicted according to the orientation difference of grains in Fig. 10. It can be seen from Fig. 10, DRX occurs at all strain rates and the size of DRX grains is larger under the strain rate of 0.001 s^{-1} than that of other strain rates. Under all the

Fig. 10 Distribution maps of different microstructures at 1200 °C under different strain rates: **a** 0.001 s^{-1} , **b** 0.01 s^{-1} , **c** 0.1 s^{-1} , **d** 1 s^{-1} , **e** 10 s^{-1}

Fig. 11 Statistical results of the microstructures in Fig. 10

Fig. 12 Distribution maps of local misorientation at 1200 °C under different strain rates: **a** 0.001 s⁻¹, **b** 0.01 s⁻¹, **c** 0.1 s⁻¹, **d** 1 s⁻¹, **e** 10 s⁻¹

deformation conditions, the deformed microstructure and the sub-microstructure are in the dominance. According to the statistic results shown in Fig. 11, the average frequency of the DRX is about 5% under all the strain rates conditions, and the frequency of the sub-structured and deformed microstructures in total is about 95%. The frequencies of DRX for all deformation conditions remain almost the same, but the frequency of sub-microstructure and deformed microstructure changes remarkably with increasing strain rate. At the lowest strain rate of 0.001 s⁻¹, the frequencies of sub-structured and deformed microstructure are basically consistent, but the frequency of deformed microstructure increases to 65% and the frequency of sub-microstructure decreases to 30% when the strain rate increases to 0.01 s⁻¹, which is mainly due to the higher strain rate and less time for DRV. As the strain rate increases further, the frequency of deformed microstructure decreases slowly, but the frequency of sub-microstructure increases gradually. This phenomenon indicates that even though higher strain rate means more deformed structure, the frequency of deformed microstructure decreases enormously as more energy is used for the microstructure evolution, which is in accordance with the results in the power dissipation maps in Fig. 7.

With increasing strain rate, more deformed microstructures transform to sub-microstructures. This result could be confirmed by the local misorientation maps in Fig. 12. Compared to the strain rate of 0.001 s⁻¹, the local orientation difference is more notable for forming more deformed microstructure at the strain rate of 0.01 s⁻¹. While the strain

rate continues to increase, the local misorientation gradually degrades for more energy dissipation in DRV.

Compared with the energy dissipation factors in the safe zone mentioned above, the values of energy dissipation factors in the instability region are less than 0.25. At the same time, it can be seen from the hot processing maps under different true strain conditions that the area occupied by unstable zone can reach 30% or even 40%. Therefore, in order to avoid the emergence of hot processing defects, and better understand the microstructure characteristics of instability region, the microstructures of the specimens deformed in the instability region are observed, as shown in Fig. 13. Compared with other alloys, hot deformation defects [48], like severe localized deformation, crack, and incomplete recrystallized grains, are not obvious in this experimental steel. Only certain deformation localization marked with white arrows in Fig. 13, which makes the grain boundary curved, occurred during hot deformation in the instability region. It could be noted that the inhomogeneity of the microstructure is the dominated feature when deformed in instability regions. More work is needed to investigate the relationship between the energy dissipation and microstructure evolution in details.

Fig. 13 SEM micrographs of the specimens deformed in the instability region at true strain of 0.6: **a** 900/10 s^{−1} °C, **b** 1000/10 s^{−1} °C

4 Conclusions

Hot working behavior of an ultra-high-strength Fe–Ni–Co-based maraging steel was investigated in the temperature range of 900–1200 °C and strain rate range of 0.001–10 s^{−1}. Based on the experimental results, the following conclusions can be drawn:

- (1) The flow stress of the experimental steel decreased with increasing temperature and increased with increasing strain rate. The precise critical value of dynamic recrystallization was determined by theoretical mathematical calculus method.
- (2) The apparent activation energy of the experimental steel was 439.311 kJ mol^{−1}. The hyperbolic sine law equation embracing the *Z* parameter was depicted as follows:

$$Z = \dot{\epsilon} \exp\left(\frac{439311}{RT}\right) = 0.9916 \times 10^{18} [\sinh(0.0069 \times \sigma)]^{4.808}.$$

- (3) The true strain has no effect on the power dissipation maps of the experimental steel when the true strain was in the scope of 0.1–0.6, while the instability zones were intensively sensitive to the true strain.
- (4) The processing maps presented large unstable zone and limited hot working parameter windows. The optimum windows were 1100–1200 °C/0.001–1 s^{−1} and 1000–1100 °C/0.001–0.1 s^{−1} with relatively high power dissipation efficiency range of 0.2–0.4.
- (5) Because of the high Mo content in the experimental steel, high dynamic activation energy, $Q = 439.311$ kJ mol^{−1}, was achieved, indicating that dynamic recrystallization is difficult to occur in the hot deformation process.

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