Hot Deformation Behavior and Processing Map of a Cu-Bearing 2205 Duplex Stainless Steel

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Abstract

The hot deformation behavior and processing map of Cu-bearing 2205 duplex stainless steel (2205-Cu DSS) were investigated at temperatures of 950–1150 °C and strain rates of 0.01–10 s⁻¹. The effects of Cu addition and different deformation parameters on deformation behavior were, respectively, characterized by analyzing flow curves, constitutive equations and microstructures. The results indicated that the shapes of flow curves strongly depended on the volume fraction of two phases. When deformed at low strain rate, DRV in ferrite was prompted with increase in the temperature and was further developed to continuous DRX. At high strain rate, flow localization preferentially occurred in ferrite at low deformation temperature due to the strain partitioning and relatively less fraction of ferrite. The activation energy for 2205-Cu DSS was 452 kJ/mol and was found to connect with the variation of strain, strain rate and deformation temperature. The optimum hot deformation parameters for 2205-Cu DSS were obtained in the temperature range of 1100–1150 °C and strain rate range of 0.1–1 s⁻¹ with a peak power dissipation efficiency of 41%. Flow localization was the main way to lead to flow instability. Meanwhile, the Cu-rich precipitates were generated within a few ferrite grains when deformed at temperature lower than 1000 °C. The interaction between dislocations and Cu-rich precipitates at high strain rate, as well as the limited DRV in ferrite and DRX in austenite, contributed to the complex microstructure and flow behavior.

Keywords Cu-bearing duplex stainless steel · Hot deformation · Processing map · Dynamic recrystallization · Dynamic recovery

1 Introduction

Duplex stainless steels (DSSs) are an important class of structural materials due to their good combination of mechanical property and corrosion resistance under critical working conditions [1–4]. Because of their excellent properties, DSSs are widely used in different industries, such as chemical tankers, pulp and paper manufacturing. As an alloying element, copper (Cu) in steels can offer numerous properties and has attracted extensive attention in recent years [5–8]. The addition of Cu in steels can not

only improve the strength but also endow superior antibacterial property for the steels. Especially for the Cu-bearing DSSs, previous studies have shown that the addition of Cu in 2205 DSS exhibited good antibacterial property and higher resistance to microbiologically influenced corrosion (MIC) [9, 10].

Although the advantageous and excellent properties DSSs possess, the hot processing and warm rolling of these materials require strict controls owing to the different deformation behaviors of the coexisted ferrite and austenite phases in DSS, which results in relatively poor hot workability. Several investigations have been carried out to understand hot deformation [11, 12] and warm-rolling [13, 14] behavior of DSS. It has been well established that the single-phase ferritic steels, with rather high stacking-fault energy (SFE), generally undergo dynamic recovery (DRV) during hot deformation [15]. Austenitic steels, on the contrary, which are characterized by a low SFE, undergo limited DRV and transform to dynamic recrystallization (DRX) when the deformation energy reaches the critical level [16–18]. Meanwhile,

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when hot deformation for DSS is performed at a constant deformation temperature and strain rate, the strain is mostly accommodated by the softer ferrite phase at the commencement of straining and DRV is the dominant deformation restoration mechanism. As deformation proceeds, the applied stress is shifted from softer ferrite to the stronger austenite at higher strain and DRX will take place [17]. In short, the co-existence of softer ferrite and stronger austenite causes strain partitioning at early deformation stage, which results in an internal stress at the ferrite/austenite boundary, and further complicates the hot deformation behavior of DSS.

In spite of the complexity for deformation behavior of DSS, a large number of researches have been made on the dynamic restoration mechanism. It is generally accepted that the dynamic restoration behaviors of austenite and ferrite in DSS are identical with single-phased austenitic and ferritic steels, with softening by DRV in ferrite and that of by DRX in austenite [19–21]. Meanwhile, Yang and Yan [22] described the DRX and DRV in DSS mainly depend on the variation of phase content, deformation temperature and strain rate, which is similar to single-phased austenitic [23] and ferritic steels [24]. However, it should be noted that the dissimilar restoration mechanism and strain partitioning during hot deformation could lead to the extremely poor hot ductility of DSS. Consequently, in order to obtain the optimum window for hot deformation and avoid the occurrence of flow instability, a practical and widely used method is the processing map. The processing map, which was first proposed by Raj [25] and based on the dynamic materials model (DMM) [26], is a powerful tool for the optimization of processing parameters and is used in various materials in a widespread manner [27, 28].

Since Cu is an austenitic element in steels and also can effectively increase the SFE of austenitic steel, the addition of Cu in DSS will certainly affect the phase volume fraction at different deformation temperatures and subsequently influence the flow behavior and dynamic restoration mechanism during the deformation. Hence, the purpose of this study is to investigate the hot deformation behavior of this novel Cu-bearing 2205 DSS (2205-Cu) by means of the hot compression test and the evaluation of strain–stress flow curves, microstructures, constitutive equations and processing maps.

2 Experimental Materials and Procedures

The chemical composition of the 2205-Cu DSS used in this study is as follows: 0.008% C, 0.52% Si, 0.95% Mn, 0.01% P, 0.003% S, 22.37% Cr, 5.43% Ni, 3.14% Mo, 0.21% N, 2.62% Cu and Fe in balance (all in wt%). The hot compression tests were performed on a Gleeble-3800 thermal/mechanical simulator at temperatures range of 950–1150 °C and strain

rates range of 0.01–10 s^{−1} with cylindrical samples 8 mm in diameter and 12 mm in height (Φ8×12 mm). Graphite foil and tantalum sheet were used as the lubricant on both contacting surfaces of specimens during hot compression. A platinum–rhodium thermocouple was used to monitor the temperature accurately. Each sample was heated to the deformation temperature at a rate of 10 °C/s. After isothermal stabilizing for 5 min at different deformation temperatures, all the samples were deformed up to 50% (true strain 0.693). After deformation, the deformed samples were water quenched immediately to preserve the microstructure. The deformation procedure is illustrated in Fig. 1.

The deformed samples were cut along the vertical axis and prepared for microstructure analysis by the standard mechanical grinding and polishing. Thereafter, electrolytic etching in 20% KOH aqueous solution was used for optical metallographic observation by a Zeiss Observer Z1 m microscope. Thin foils with 3 mm in diameter for transmission electron microscope (TEM, JEM 2100) investigation were mechanically ground to 50 μm in thickness, further electrolytic polished in a solution of 10 vol% HClO₄ ethanol by a Struers Tenupol-3 twin-jet electropolisher. The X-ray diffraction (XRD) measurement was conducted on a D/Max-2500PC X-ray diffractometer using Cu-Kα incident radiation in a 2 Theta range of 40°–100° at scanning rate of 4°/min.

3 Results

3.1 Initial Microstructure of 2205-Cu Duplex Stainless Steel

The initial microstructure and XRD patterns of the 2205-Cu DSS are shown in Fig. 2. As displayed in Fig. 2, the microstructure of the as-received 2205-Cu DSS is composed of austenite (bright) and ferrite (dark) phases, each phase paralleling to the rolling direction. The strip-like

Fig. 1 Deformation procedure for the experimental 2205-Cu DSS

Fig. 2 Initial microstructure **a**, XRD patterns **b** of as-received 2205-Cu DSS before hot deformation. The inset in **a** shows the magnification of initial microstructure

Fig. 3 Volume fraction of austenite in 2205-Cu DSS as a function of deformation temperature

austenite is embedded in the ferrite matrix. Figure 3 shows the volume fraction of austenite as a function of deformation temperature, which was measured by microstructure analysis. As it can be seen, the austenite volume fraction decreases from $59 \pm 1.3\%$ at 950 °C to $51 \pm 1.8\%$ at 1050 °C and down to $43 \pm 1.2\%$ at 1150 °C. It should be noted that the austenite volume fraction is much higher than that of the previous study at the same deformation temperature [29]. This may be attributed to the addition of Cu, an austenite stabilizer, increasing the austenite stability in DSS at high deformation temperature.

3.2 Flow Stress–Strain Curves

The flow stress–strain curves of 2205-Cu DSS deformed at different deformation conditions are shown in Fig. 4. Obviously, the flow curves are very much sensitively dependent on the variation of deformation parameters, i.e., temperature and strain rate. The flow stress evidently increases with increase in the strain rate and decreases with increase in the temperature. Meanwhile, the flow curve remarkably increases at the beginning of deformation, ascribed to the work hardening effect and further reaches to a relatively steady state [30].

As is well known, the shape of flow curves can reflect the microstructure change and interaction between work hardening and dynamic restoration mechanism. Dynamic recrystallization (DRX) and dynamic recovery (DRV) are the two main restoration mechanisms for steels. While the dominant softening mechanism during hot deformation depends on the Zener–Hollomon (Z) value, which embodied the incorporated effect of deformation temperature and strain [31].

Two main features of the flow curves in the present study should be noted. On the one hand, at low deformation temperature (less than 1100 °C), the flow curve exhibits faint peak-type shape which may indicate the taking place of DRX, as shown in Fig. 4a–c. With increase in the temperature, the peak becomes broad and even disappears at high deformation temperature, showing a flat-top shape curve without clear peaks. This kind of flow curve is generally considered as the DRV occurring during the deformation. On the other hand, at low strain rate (less than 1 s^{-1}), the flow curve is nearly a typical DRV shape, as shown in Fig. 4d. With increase in the strain rate, the flow curve develops to DRX type curve. Besides, some of the flow curves even demonstrate multiple peaks, e.g., $1000 \text{ °C}/10 \text{ s}^{-1}$. The

Fig. 4 Flow stress–strain curves of the 2205-Cu DSS deformed at different deformation conditions

relationship between flow curve characteristics and restoration mechanism will be discussed in detail in next section combined with the microstructure evolution.

3.3 Constitutive Equations Analysis

Constitutive equations are commonly used to reflect the relationship among flow stress, temperature and strain rate during hot deformation. The dependence of flow stress on the deformation temperature and the strain rate can be analyzed by the well-known hyperbolic sine function [32], which can be applied for a wide range of temperatures and strain rates. The hyperbolic sine function is given as follows:

$$Z = \dot{\epsilon} \exp\left(\frac{Q}{RT}\right) = A[\sinh(\alpha \cdot \sigma)]^n, \quad (1)$$

where Z stands for the Zener–Hollomon parameter, Q accounts for the activation energy, σ denotes the peak flow stress. A , α and n are material constants, which can be

calculated using the regression fitting of logarithmic form of Eq. (1) as described elsewhere [5].

Figure 5 illustrates the regression analysis of the relationship among flow stress, strain rate and deformation temperature for the 2205-Cu DSS. As shown in Fig. 5, by applying the linear regression, the material constants can be derived from the average slope of linear regression lines. The values of A , α , n and Q in the present study are 7.852×10^{16} , 0.00939 MPa^{-1} , 4.495 and 452 kJ/mol, respectively. By replacing the obtained material constants into Eq. (1), the constitutive equation containing Z parameter is developed as follows:

$$Z = \dot{\epsilon} \exp\left(\frac{452,000}{RT}\right) = 7.852 \times 10^{16} [\sinh(0.00939 \cdot \sigma)]^{4.495}. \quad (2)$$

It is worthy to note that the activation energy for 2205-Cu DSS is 452 kJ/mol, and the relationship between flow stress and Z parameter is well fitted linearly on the base of this activation energy. For duplex stainless steel, two phases

Fig. 5 Regression fitting of the hyperbolic sine function for 2205-Cu DSS under the different flow stress, strain rate and deformation temperature

have a significant difference in activation energy. Generally, the hot deformation activation energies for austenite steels (390–500 kJ/mol) are larger than that of ferrite steels (270–400 kJ/mol), implying that the flow stress for austenite steel is more temperature dependent [33, 34]. Besides, the temperature range and deformation type, e.g., hot compression or torsion, can also significantly affect the value of activation energy.

Figure 6 exhibits the dependence of flow stress with Z parameter under different deformation conditions. It can be clearly seen that the activation energy of 2205-Cu DSS at high temperature (1050–1150 °C) is higher than that of at low temperature (950–1050 °C), i.e., 519 kJ/mol and 421 kJ/mol, respectively. This result can be attributed to the different volume fractions of two phases at different deformation temperatures. In the case of DSS, it has been proved that the overall activation energy is a kind of weighed average of each phase, depending on the volume fraction of two phases [35]. In the present study, the volume fraction of

Fig. 6 Dependence of flow stress with Z parameter under different deformation conditions according to hyperbolic sine law

softer ferrite increases with increase in the deformation temperature, as shown in Fig. 3. Since the activation energy of ferrite phase is lower than that of austenite at the same deformation temperature, thus it is easy to deduce that the increase in ferrite phase, in addition to the softening effect of high temperature, will lead to the decrease in the activation energy at high deformation temperature.

Except for the effect of temperature on the activation energy, it should be noted that Eq. 1 is given at constant strain, implying that activation energy is strain dependent. Figure 7 shows the variation of activation energy with strain and strain rate. Clearly, the activation increases monotonously with increase in the strain and strain rate. This is mainly due to the strain partitioning caused by the co-existence of softer ferrite and stronger austenite in DSS. At initial stage of deformation, the strain is mostly accommodated by the softer ferrite. While with increase in the deformation strain, the stronger austenite starts to deform and thus increasing the activation energy. As for the effect of strain rate, the austenite deformation is prone to be triggered at high strain rate and likewise, leading to the increase in the activation energy.

3.4 Establishment of Processing Map

The processing maps including power dissipation map and instability map are based on the DMM model which has been described by Prasad [26]. In the DMM model, the workpiece is regarded as a power dissipater. The parameter η , known as efficiency of power dissipation, indicates how efficient the energy dissipates by microstructure change and hence can be considered as the exhibition of hot workability. The η is expressed as:

$$\eta = \frac{2m}{m+1}, \quad (3)$$

where m is the strain rate sensitivity. Besides, in order to detect the deformation instability region, the flow instability criterion, ξ , based on the continuum principles proposed by Ziegler [36], is used in the present study and given as:

$$\xi = \frac{\partial \ln[m/(m+1)]}{\partial \ln} + m \leq 0. \quad (4)$$

According to the Ziegler criterion, a negative ξ value means the occurrence of flow instability.

Figure 8 shows the processing map of 2205-Cu DSS which was obtained on the basis of flow stress data at a true strain of 0.6. The contour numbers inside the map represent the values of power dissipation efficiency, and shaded area represents the unstable hot deformation regime. As depicted in Fig. 8, the map shows two domains with peak power efficiency of 37% occurring at 1050 °C and 0.1 s⁻¹ (domain I) and 41% occurring at 1100 °C and 0.1 s⁻¹ (domain II), which is considered as the process window for hot working. Figure 9a, b gives the detailed microstructure of 2205-Cu DSS corresponding to the deformation conditions marked domain I and II on the processing map, respectively. In domain I where the power efficiency is 37%, as shown in Fig. 9a, the microstructure indicates limited DRX in austenite and extended DRV or the so-called continuous DRX in the ferrite [37, 38]. Grain boundaries can be seen in the austenite island, and fine ferrite grains/subgrains are generated in the ferrite with size of ~5 μm in diameter. Besides, two phases are aligned in the rolling direction of the previous deformation process. When the power efficiency is 41%, the degree of DRX both in austenite and ferrite becomes larger with increase in the temperature as shown in Fig. 9b. The size of newly formed ferrite grains increases to ~10 μm.

Fig. 7 Activation energy as a function of strain and strain rate

Fig. 8 Hot processing map for 2205-Cu DSS at strain of 0.6

Fig. 9 Optical micrographs of 2205-Cu DSS obtained at 1050 °C/0.1 s⁻¹ **a**, 1100 °C/0.1 s⁻¹ **b**, 1000 °C/10 s⁻¹ **c**, 1100 °C/10 s⁻¹ **d**

Meanwhile, partial DRX austenite grains are no longer clung to the austenite island, but are percolated and dispersed into the ferrite region. In this region, in spite of the existence of strain partitioning during deformation, deformed at intermediate temperature and strain rate can possess proper volume fraction of two phases and further promote the appearance of DRX in two phases, leading to a safe hot deformation domain.

From the processing map, it can be clearly seen that the unstable domain for hot working occurs at high strain rate (≥ 1 s⁻¹) and all the deformation temperatures, e.g., 1000 °C and 10 s⁻¹ (domain III), and 1100 °C and 10 s⁻¹ (domain IV). Figure 9c, d shows the microstructure of the specimens after deformed at 1000 °C/10 s⁻¹ and 1100 °C/10 s⁻¹, corresponding to domain III and domain IV, respectively. No crack, wedge or macro-cavity can be found in the microstructures in the flow instability region. Flow localization is the main way to lead to the flow instability. During the high strain rate deformation, a great mass of plastic deformation is concentrated on softer ferrite region at early deformation stage within a very short time and thus causing the flow localization in the ferrite region. After then, the stronger austenite is triggered to deform at late stage of deformation.

The DRX grains/subgrains between the wavy or serrated austenite grain boundaries, marked by white arrow, can be clearly found in Fig. 9c, d.

According to the processing map and microstructure discussed above, the optimum hot deformation parameters for 2205-Cu DSS can be worked out to be in the temperature range of 1100–1150 °C and strain rate range of 0.1–1 s⁻¹ with a peak power dissipation efficiency of 41%.

4 Discussion

As mentioned above, flow curves and deformation microstructures can reflect the softening mechanism during hot deformation. Generally, DRX and DRV are two major softening mechanisms for metallic materials deformed at high temperature. In this section, the microstructures of 2205-Cu DSS at different deformation conditions are chosen to analyze the influence of strain rate and temperature on the softening mechanisms for both ferrite and austenite.

Figure 10 shows the microstructures of 2205-Cu DSS deformed at different temperature and strain rate. Obviously, the volume fraction of austenite decreases as the

Fig. 10 Microstructure for 2205-Cu DSS obtained at 950 °C/0.01 s⁻¹ **a**, 1050 °C/0.01 s⁻¹ **b**, 1150 °C/0.01 s⁻¹ **c**, 950 °C/10 s⁻¹ **d**, 1050 °C/10 s⁻¹ **e**, 1150 °C/10 s⁻¹ **f**. Hereinto, black arrows denote the DRV in ferrite grains, and white arrows represent the DRX in austenite grain

deformation temperature rises, which is consistent with the results shown in Fig. 3. The effect of strain rate on the microstructure of deformed samples is shown in Fig. 10a, d. At low temperature and low strain rate (950 °C/0.01 s⁻¹), the deformed microstructure shows some weak grain boundaries in the ferrite, indicated by black arrows, as well as the bulging of serrated original austenite grain boundaries, as shown in Fig. 10a. It can be deduced that deformation at low temperature and strain rate may lead to limited DRV in ferrite while plenty serrated austenite grain boundaries result in the beginning of DRX in austenite. With increase in the strain rate (950 °C/10 s⁻¹), the degree of local bulging of serrated grain boundaries in austenite increases, indicating that DRX is accelerated. Meanwhile, the flow localization is also observed in the ferrite region. This may be due to the plenty austenite grain boundaries can offer abundant nucleation site for DRX grains. While high strain rate makes the softer ferrite to be deformed rapidly, which leads to the flow localization in ferrite and prompts the stronger austenite to be triggered easily.

Figure 10c exhibits the microstructure of 2205-Cu DSS deformed at high temperature and low strain rate (1150 °C/0.01 s⁻¹). Distinct grain boundaries in ferrite (marked by black arrows) and grains/subgrains in austenite (marked by white arrows) can be observed, indicating well-developed DRV in ferrite and a certain content of DRX in austenite. In spite of the fact that deformed at high temperature can promote the occurrence of DRX, the lower austenite

volume fraction at high temperature restricts the nucleation and growth in the less interior austenite boundaries. The microstructure deformed at high temperature and high strain rate (1150 °C/10 s⁻¹) is shown in Fig. 10f. Sharp boundaries and subgrains in ferrite indicate the extended DRV in ferrite, while few visible austenite boundaries imply the suppression of DRX. In comparison with Fig. 10c, high strain rate promotes the degree of DRV in ferrite and even leads to the flow localization in partial region of ferrite.

Figure 10b, e shows the microstructure of 2205-Cu DSS obtained at the intermediate temperature. Obviously, sharper boundaries in two phases manifest the development of DRV and DRX. Although ferrite, characterized by relatively high SFE, is apt to be softened by DRV, it is possible to create a deformation condition that restricts DRV and prompts DRX on reaching a certain level of strain. In the present study, few continuous DRX ferrite grains occurred owing to the inhomogeneous deformation in ferrite at high strain rate, as shown in Fig. 10e. This result is consistent with the previous reports that DRX was developed in ferritic steels after deformed at high strain rate and high temperature [39, 40]. Considering deformation at high or low temperature, the difference in softening mechanism can be inferred that the volume fractions of ferrite and austenite are similar at the intermediate temperature.

To further analyze the deformation behavior of 2205-Cu DSS, Fig. 11 shows the TEM images of the deformed specimens at different deformation temperature and strain rate.

Fig. 11 TEM images of the 2205-Cu DSS under different deformation conditions: **a** 950 °C/0.01 s⁻¹, **b** 950 °C/10 s⁻¹, **c** 1150 °C/0.01 s⁻¹, **d** 1150 °C/10 s⁻¹

Obviously, a complex and inhomogeneous microstructure can be found in both ferrite and austenite regions. The strain is accommodated by ferrite with mass of dislocations within the original ferrite grains, as shown in Fig. 11a. Besides, in certain area of the ferrite, the grains adopt a bamboo type of microstructure. When deformed at high strain rate and low temperature, the flow localization with a mass of tangled dislocations is easily seen in ferrite region. Along with further increase in the deformation temperature, the ferrite exhibits a well-developed recovered structure and is extended to continuous DRX, as can be seen in Fig. 11c. Meanwhile, except for the undeformed austenite grain, limited DRX with visible grain boundary in austenite can also be observed at high strain rate.

Through above analysis, it can be inferred that the complexity of hot deformation behavior of 2205-Cu DSS at different temperature and strain rate is mainly caused by the volume fraction variation of ferrite and austenite. Flow localization is apt to occur in ferrite at low deformation temperature and high strain rate due to the less fraction of ferrite. Whereas DRX in austenite is easy to be triggered at high strain rate owing to the rapid deformation. As for the low strain rate, DRV in ferrite accelerates with increase in the temperature and is extended to continuous DRX. While DRX in austenite at high temperature is suppressed due to

the low volume fraction of austenite and the strain accommodation by ferrite. At intermediate temperature of 1050 °C, DRX and DRV in two phases are well-developed resulting in the generation of new substructure in two phases.

Besides, the addition of Cu also has a significant effect on the hot deformation behavior of duplex stainless steel. As well known, Cu in steel is an austenitic element and is known to increase the SFE and tempering resistance [41]. For the DSS, the distributions of Cu in two phases are also different which may lead to the variation of deformation mechanism resulting from the change of SFE. Figure 12 exhibits SEM-EDS results of ferrite and austenite for 2205-Cu DSS at deformation temperature of 1000 °C. As clearly seen, the contents of Cr, Cu, Ni and Mo in ferrite and austenite are different, and the contents of Cu and Ni in austenite are higher than those in ferrite. According to the previous study [11], DRX in austenite is delayed due to the strain partitioning in DSS. Nonetheless, the increment of SFE by the higher contents of Cu and Ni in austenite may be another factor to restrict the development of DRX in austenite during hot deformation.

Furthermore, because of the limited solid solubility of Cu in ferrite and austenite, Cu-rich phase will precipitate out from the steel matrix at appropriate temperature for proper time. For example, some recent studies analyzed the Cu-rich

Fig. 12 SEM-EDS analysis of austenite and ferrite for 2205-Cu DSS at temperature of 1000 °C

precipitation behavior in single-phase austenitic steel and found that the formation of Cu-rich precipitates occurred immediately after an appropriate aging treatment and the interface between Cu-rich phase and austenitic matrix remained coherent during the aging process [42, 43]. For the Fe-Cu based ferritic steel, lots of studies suggested that Cu-rich precipitates remained coherent relation with the body-centered cubic α -Fe matrix at an early stage of aging treatment and transformed into a twinned 9R structure with an increase in the aging time [44, 45].

In the present study, Cu-rich precipitates can also be found in the matrix of 2205-Cu DSS. Figure 13 presents the TEM images of Cu-rich precipitates after deformed at 950 °C/10 s⁻¹ and 1000 °C/10 s⁻¹. As clearly seen, Cu-rich precipitates are precipitated out within the ferrite grains, with size of about 15 nm in diameter. No precipitation is observed when the specimens are aged above the temperature of 1000 °C. These Cu-rich precipitates may be generated during the heating process and remained undissolved at deformation temperature (lower than 1000 °C) or

precipitated from matrix at deformation temperature. But, either way, the existence of Cu-rich precipitates can no doubt influence the softening mechanism of 2205-Cu DSS through the interaction with dislocations. During deformation, the Cu-rich precipitates can pin the dislocations and increase the deformation resistance at the deformation temperature [46]. The interaction between glide dislocations and Cu-rich precipitates will make the slip of dislocations difficult, therefore retard the DRV in ferrite by restricting the annihilation and rearrangement of glide dislocations into the low-energy walls, as can be seen in Fig. 13. In addition, the interaction between dislocations and Cu-rich precipitates can also affect the shape of flow curves. As shown in Fig. 4d, the flow curves demonstrate fluctuation character with multiple peaks, which may result from the complex softening mechanism during deformation. When the specimens are deformed at high strain rate and low temperature, e.g., 950 °C/10 s⁻¹, the softening mechanism is extremely complex, consisting of limited DRV in ferrite, DRX in austenite and interaction between dislocations and Cu-rich precipitates. This

Fig. 13 TEM bright field images of Cu-rich precipitates in 2205-Cu DSS under different deformation conditions: **a** 950 °C/10 s⁻¹, **b** 1000 °C/10 s⁻¹. The inset is the corresponding electron diffraction patterns of the matrix

synergistic effect of DRX, DRV and dislocations interaction leads to a complicated deformation microstructure and flow curves as well.

5 Conclusions

Hot deformation behavior of a Cu-bearing 2205-Cu DSS was studied in wide temperature and strain rate range by using the hot compression test. Based on the experimental results, the following conclusions can be drawn:

1. The deformation behavior of 2205-Cu DSS at elevated temperatures strongly depended on the phase volume fraction. When deformed at high strain rate, flow localization was apt to occur in ferrite at low deformation temperature due to the less fraction of ferrite. At low strain rate, DRV in ferrite was accelerated with increase in the temperature and was extended to continuous DRX. While DRX in austenite at high deformation temperature was postponed by the strain accommodation in ferrite and few internal boundaries in austenite.
2. The activation energy of 2205-Cu DSS for hot deformation was 452 kJ/mol, which was found to increase with increase in strain rate and strain and decrease with increase in temperature. This was mainly due to the volume fraction variation of stronger austenite and strain partitioning at different temperature, strain rate and strain.
3. The optimum window for hot deformation of 2205-Cu DSS could be worked out to be in the temperature range of 1100–1150 °C and strain rate range of 0.1–1 s⁻¹ with a peak power dissipation efficiency of 41%. The constitutive equation containing Z parameter was obtained and described as follows:

$$Z = \dot{\epsilon} \exp \left(\frac{452,000}{RT} \right) \\ = 7.852 \times 10^{16} [\sinh (0.00939 \cdot \sigma)]^{4.495}.$$

4. The Cu-rich precipitates were generated within a few ferrite grains when deformation temperature was lower than 1000 °C and further complicated the hot deformation behavior of 2205-Cu DSS. The interaction between dislocations and Cu-rich precipitates during high strain rate, as well as the limited DRV in ferrite and DRX in austenite, contributed to the complex microstructure and flow behavior of 2205-Cu DSS.

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