Hot deformation behavior and microstructural evolution of 2205 duplex stainless steel

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Abstract

The equiaxed compression test was used to study the microstructure evolution and the dynamic recrystallization behavior of 2205 duplex stainless steel during hot deformation under 850–1100 °C with strain rates of 0.01 s\(^{-1}\), 0.1 s\(^{-1}\), 1 s\(^{-1}\) and 10 s\(^{-1}\). The calculated values of thermal deformation activation energy \(Q\) and stress index \(n\) were 351.58 kJ mol\(^{-1}\) and 3.8543, respectively. On this basis, the Arrhenius type constitutive equation was successfully established, the microstructure of the sample was studied by electron backscatter diffraction (EBSD), and the changes of the ferrite and austenite phase under different deformation parameters were analyzed. It was seen from the phase boundary and grain boundary diagrams that, at the same temperature, as the deformation rate increased, the austenite phase content slowly decreased, while the ferrite phase content increased. The deformation amount of the two phases under the same strain at different temperatures and strain rates were respectively obtained from the grain orientation scatter diagram and the changes of recrystallization, deformation and substructure under different deformation parameters were obtained from the recrystallization diagram.

1. Introduction

Duplex stainless steel refers to the steel containing both austenite and ferrite, which is a type of stainless steel with a two-phase structure of \(\alpha\) and \(\gamma\). Therefore, it owns the characteristics of both austenitic stainless steel and ferritic stainless steel. Compared with ferritic stainless steel, duplex stainless steel not only retains the advantages of low linear expansion coefficient and high thermal conductivity of ferritic stainless steel, but also has high toughness, low brittleness transition temperature, which significantly improves its welding performance and resistance to intergranular corrosion. Compared with austenitic stainless steel, duplex stainless steel has higher strength, and better resistance to stress corrosion cracking and pitting corrosion. Consequently, it is widely used in petrochemical, paper, and other industries \cite{1–4}.

The austenite and ferrite phases in duplex stainless steels differ greatly in terms of composition, microstructure and mechanical properties. The ferrite phase has lower thermal strength and higher stacking fault energy, making it more prone to dislocation bunching and dynamic recovery during thermal deformation. In contrast, the austenite phase has higher thermal strength, and lower stacking fault energy during thermal deformation, which will make dislocations more prone to entanglement, resulting in processes such as the dynamic recovery often being suppressed \cite{5–8}. The difference in two-phase mechanical properties and deformation behavior in duplex stainless steel makes its plastic deformation behavior more complex, and the difficulty of alloy hot working correspondingly increase \cite{9}, which results in the generally poor surface quality of the finished product. In recent years, more and more attention has been paid to the study of hot working properties and plastic deformation behavior of duplex stainless steel. Previous studies \cite{10–12} have shown that the high-temperature plastic deformation of the alloy first occurs in the soft ferrite, and the deformation is controlled by the ferrite phase under the small stress variable; then
with the increasing of the strain, the stress in the ferrite gradually increases, causing the deformation dislocation to gradually transfer into the austenite from the two-phase interface. In addition, the distribution of strain in austenite and ferrite phases will be affected by the deformation conditions. Studies \[13, 14\] have shown that the higher temperature and the lower strain rate have positive effect on the plastic deformation of the ferrite phase during hot compression of duplex stainless steel 2205; while the lower deformation temperature and the higher strain rate are beneficial to the plastic deformation of the austenite phase.

The properties of materials depend on the internal structure, and the different crystal orientation distribution can also reflect the final properties of materials. In recent years, the EBSD technology, overcoming the problems of small analysis region and poor statistics of TEM, has been widely used in the characterization of material microstructure \[15–18\].

In this paper, the microstructure evolution of 2205 duplex stainless steel during hot deformation at different temperature and strain rates was studied, the microstructure information such as the proportion of different phases, morphology, and the degree of recrystallization was analyzed by EBSD, and the constitutive model was established.

## 2. Materials and methods of test

Table 1 shows the chemical composition of 2205 duplex stainless steel used in this study where the initial proportions of ferrite and austenite are about 57.52% and 42.48%, respectively. An isothermal uniaxial compression test was carried out on a cylinder with a diameter of 10 mm and a height of 15 mm using Gleeble-3800 thermal simulator. Figure 1 shows the detailed experimental procedure of 2205 duplex stainless steel investigated. The sample was heated to 1230 °C at a heating rate of 5 °C s\(^{-1}\) for 180 s, then cooled to a different deformation temperature at a cooling rate of 5 °C s\(^{-1}\) for 30 s for uniform organization, and then subjected to compression deformation. The deformation temperature \(T\) is 850 °C, 900 °C, 950 °C, 1000 °C, 1050 °C and 1100 °C, the deformation rate is 0.01 s\(^{-1}\), 0.1 s\(^{-1}\), 1 s\(^{-1}\) and 10 s\(^{-1}\), and the true strain is 0.92. After being loaded at different strain rates, the samples were immediately treated with water cooling to retain the microstructure. The true stress-true plastic strain curve was obtained by subtracting the elastic region from the true stress-strain curve. Then, the samples were cut along the compression direction to analyze the deformed microstructure.

After high temperature compression, half of the cut samples were used to prepare an EBSD research sample. The cross sections of the thermally deformed samples were polished using standard metallographic procedures. The electrolysis was performed in a solution of 20:1 ethanol and perchloric acid. The electrolysis voltage was 35 V and the current was 1 A. A small region was taken from the uniform deformation region of the compressed sample for EBSD analysis.

### Table 1. Chemical composition of 2205 duplex stainless steel.

| C  | Cr  | Mo  | Ni  | Si  | S   | Mn  | P    | N    | Fe   |
|----|-----|-----|-----|-----|-----|-----|------|------|------|
| 0.025 | 21.83 | 3.09 | 5.45 | 0.58 | 0.003 | 1.13 | 0.024 | 0.1699 | Bal  |

![Figure 1](image-url) Figure 1. The detailed experimental procedure for the investigated 2205 duplex stainless steel investigated.
3. Analysis and discussion

3.1. Stress strain curve

Figure 2 shows the true stress-strain curve of 2205 duplex stainless steel with a strain rate of 0.01 – 10 s\(^{-1}\) in the temperature range of 850 – 1100 °C. It can be seen that the flow stress decreases with the increasing temperature and the decreasing strain rate, showing the significant effects of temperatures and strain rates on the flow stress. In the early stage of deformation, the flow stress increases rapidly with increasing strain, which is caused by the generation of dislocations and work hardening. After the rapid increase, the flow stress starts to increase slowly and after reaching a peak, tends to decrease or maintain a stable state, indicating that the dynamic softening caused by dynamic recovery (DRV) and dynamic recrystallization (DRX) is sufficient to offset work hardening. In general, DRV keeps the flow stress curve straight, while DRX makes it drop. As shown in figure 2, it is difficult to provide enough energy for the grain boundary migration at a low deformation temperature and the stress-strain curves with the strain rate of 0.01 – 1 s\(^{-1}\) show a downward trend, while the high strain rate cannot provide enough time for nucleation and growth of dynamic recrystallization grains, and the stress-strain curves with the strain rate of 10 s\(^{-1}\) show an upward trend.

At a low strain rate of 0.01 s\(^{-1}\), after the peak of the stress-strain curve appeared, the DRX softening behavior decreased significantly, and as the temperature increased, the true strain corresponding to the peak stress continued to decrease, from 0.33 to about 0.05. This is because as the temperature increases, the atom diffusion ability increases, the grain boundary migration ability increases, and the activation energy for recrystallization occurs decreases. With the increase of the strain rate and the shortening of the thermal compression deformation time, the DRX softening behavior of the apparent decrease in the stress-strain curve is lessor even not, which shows a DRV or a form of stress rising again (figure 2(d), 10 s\(^{-1}\) strain rate). This is due to the shortening of deformation time and insufficient incubation period of recrystallization nucleation, which can not effectively eliminate the distortion caused by dislocation slip. As the strain rate increases to 10 s\(^{-1}\), the deformation time is shortened, and the DRV can not completely eliminate the distortion caused by dislocation slip, so that the curve shows that the stress rises again. In addition, at 850 °C, at 0.01 and 0.1 s\(^{-1}\), there is a significant yield plateau on the curve. The yield plateau is a kind of yield behavior of low-carbon steel and low-strength low-alloy steel, which is mainly caused by the ‘drag’ slip displacement of carbon and nitrogen atomic gas groups (Coriolis atmosphere). The results also show that at low temperature and low strain rate, the bcc phase in 2205 dual phase steel first slips and deforms, and then 3.3 microstructure analysis will further explain the different microstructure evolution behavior of bcc phase and fcc phase in 2205 dual phase steel.
3.2. Dynamics analysis

The relationship between the deformation conditions of the material during the thermal deformation process is usually used to predict the forming process of the material. The flow stress equation can describe the relationship between the deformation conditions of the material during the thermal deformation, which is usually used to predict the forming process of the material. In thermal deformation, the flow stress $\sigma$ is related to the deformation temperature $T$ and the strain rate $\dot{\varepsilon}$.

$$\dot{\varepsilon} = A_1 \cdot \sigma^\alpha \exp[-Q/(RT)] \quad \alpha \sigma < 0.8$$

$$\dot{\varepsilon} = A_2 \cdot \exp(\beta \alpha) \exp[-Q/(RT)] \quad \alpha \sigma > 1.2$$

$$\dot{\varepsilon} = A \cdot [\sinh(\alpha \sigma)]^n \exp[-Q/(RT)] \quad \alpha \sigma > 1$$

Where $\dot{\varepsilon}$ is the strain rate, $\sigma$ is the peak stress, $Q$ is the activation energy for deformation, $R$ is the gas constant (8.3145 kJ mol$^{-1}$), $T^{-1}$ is the deformation temperature (K), $A_1, A_2, A, n,$ and $\beta$ are material constants, and $\alpha$ is a stress multiplier and can be described as $\alpha = \beta/\eta_i$.

The power function (equation (1)) is used for lower stress values compared with the exponential function (equation (2)), whereas the hyperbolic sine function (equation (3)) can be applied to the whole stress range. Therefore, the hyperbolic equation (equation (3)) is used to establish the constitutive relationship of 2205 duplex stainless steel.

The strain rate of a material under high-temperature plastic deformation is controlled by a thermal activation process. The relationship of flow stress $\sigma$, strain rate $\dot{\varepsilon}$, and deformation temperature $T$ is as follows:

$$Z = \dot{\varepsilon} \exp(Q/RT) = A[\sinh(\alpha \sigma)]^n$$

Where $Z$ is the Zener-Hollomon parameter, which is the strain rate factor for temperature compensation.

Take the logarithms of both sides of equation (1) and formula equation (2) to get:

$$\ln \dot{\varepsilon} = m_1 \ln \sigma + \ln A_1 - Q/RT$$

$$\ln \dot{\varepsilon} = \beta \sigma + \ln A_2 - Q/RT$$

Substituting the peak stress $\sigma$ under different thermal deformation conditions into equations (5) and (6), the linear fitting can obtain the $\ln \dot{\varepsilon} - \ln \sigma$ and $\ln \dot{\varepsilon} - \sigma$ curve under different deformation conditions as shown in figure 3, and the average slope value of the curve is $n_1 = 5.3548$, $\beta = 0.03456$, $\alpha = \beta/\eta_1 = 0.00645$.

The logarithm of the two sides of equation (3) can be obtained:

$$\ln \dot{\varepsilon} = n_{\ln} \ln \sigma + \ln A - Q/RT$$

Bringing the peak stress $\sigma$ into the equation (7), the $\ln \dot{\varepsilon} - \ln[\sinh(\alpha \sigma)]$ relationship curve under different deformation conditions obtained by linear fitting is shown in figure 4(a), and the average slope is $n = 3.8543$.

When the strain rate is constant, assuming that the deformation activation energy is independent of the temperature, the logarithm of equation (3) is arranged as:

$$\ln[\sinh(\alpha \sigma)] = \frac{\ln A - \ln \dot{\varepsilon}}{n} + \frac{Q}{nRT} \frac{1}{nRT}$$

The linear relationship of $\ln[\sinh(\alpha \sigma)] - 1/T$ of 2205 duplex stainless steel under different deformation conditions is shown in figure 4(b), and the average value of its slope is $Q/nR = 10,971.1346$. 

![Figure 3](image-url-3.png)

3.2. Linear relationship at different temperatures. (a) $\ln \dot{\varepsilon} - \ln \sigma$ (b) $\ln \dot{\varepsilon} - \sigma$. 

![Figure 4](image-url-4.png)

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Taking the average value of \( \ln \varepsilon - \ln \sinh (\alpha \sigma) \) slope and molar gas constant \( R \) into \( Q/nR = 10971.1346 \), the value of \( Q = 351.58 \) kJ mol\(^{-1}\) is within the reported apparent activation energy for low and high temperatures reported for as-received 2205 duplex stainless steel (554 kJ mol\(^{-1}\) and 310 kJ mol\(^{-1}\)) [6].

Take the logarithm of both sides of equation (4) at the same time to get:

\[
\ln Z = \ln A + n \ln \sinh (\alpha \sigma)
\]  
(9)

Substitute the deformation rate and the thermal deformation activation energy at different temperatures into equation (4) to obtain the \( Z \) value. Draw the fitting curve \( \ln Z - \ln \sinh (\alpha \sigma) \) according to equation (9) as shown in figure 4(c):

Bringing the above parameters into equation (3), the constitutive equation of 2205 duplex stainless steel is obtained as follows:

\[
\varepsilon = 1.18 \times 10^{10} \sinh (0.00645 \sigma) \times 10^{3.8543} \exp (-351580/RT)
\]  
(10)

3.3. Microstructure analysis

Figure 5 shows the different morphological characteristics of the two phases at different strain rates at 850 °C. The gray region is bcc, the yellow region is fcc, and the red line is the phase boundary. The black line corresponds to low-angle grain boundaries with misorientation between 2° and 15°. It can be seen from figure 7 that at the strain rate of 0.01 s\(^{-1}\), the austenite phase appears elliptical at the rate of 0.01 s\(^{-1}\). As the deformation rate increases, the austenite phase appears elliptical, and there are more subgrain boundaries in the austenite phase than in the ferrite phase. As the deformation rate increases to 10 s\(^{-1}\), the austenite phase is severely stretched and deformed, with a large number of black subgrain boundaries and grain boundaries appearing.

Figure 6 shows the different morphological characteristics of the two phases at different strain rates at 1000 °C. It can be seen from figure 6 that at the strain rate of 0.01, the austenite phase appears elliptical, and there are more subgrain boundaries in the austenite phase than in the ferrite phase. As the deformation rate increases to 0.1 s\(^{-1}\), the austenite phase shows no evident changes. When the deformation rate increases to 1 s\(^{-1}\), the large austenite phase is flattened, showing a tendency of fragmentation. When the deformation rate increases to 10 s\(^{-1}\), the austenite phase is severely stretched and deformed, with a large number of black subgrain boundaries and grain boundaries appearing.

Figure 7 shows the different morphological characteristics of the two phases at different strain rates at 1050 °C. It can be seen from figure 7 that at the strain rate of 0.01, the austenite phase appears elliptical, and the subgrain boundaries in the austenite phase and the ferrite phase are not obvious. As the deformation rate increases to 10 s\(^{-1}\), the austenite phase is severely elongated and deformed, with the black subgrain boundaries and grain boundaries increasing. At both rates, the ferrite phase is coarser.

It can be seen from figure 8 that at the same temperature, as the deformation rate increases, the austenite phase content decreases slowly, while the ferrite phase content increases. At 850 °C with a strain rate of 0.01 s\(^{-1}\), the proportions of the bcc phase and the fcc phase are 69% and 28%, respectively and at 1050 °C with a strain rate of 10 s\(^{-1}\), the proportions of the two phases are 65% and 35%, respectively. With the increase of the strain rate, the transformation from austenite to ferrite is accelerated. At 850 °C with a strain rate of 0.1 s\(^{-1}\), the content
of bcc phase decreases while that of fcc phase increases, which may be caused by the less conversion of fcc to bcc at this point.

Figure 9 shows the grain orientation spread (GOS) of the two phases at different strain rates at 850 °C. The strain value of the grains is determined by measuring the degree of lattice rotation. The gray region is fcc, the blue-green region is bcc, and the red line is the phase boundary. The black fine line corresponds to low-angle grain boundaries with misorientation between 2° and 15° while the black thick line represents high-angle grain boundaries with misorientation over 15°.
grain boundaries with misorientation between 2° and 15° while the black thick line represents high-angle grain boundaries with misorientation over 15°. According to the scale in figure 9, bcc colors the grains with higher strain from blue to red, and fcc colors the grains with higher strain from black to white. It can be seen from figure 9 that at the strain rate 0.01, the strain values of fcc and bcc phases are gray and blue-green, respectively. The fcc phase with high strain values shows more white regions, while the bcc phase with high strain values shows less red regions, indicating more high-strain grains in the fcc phase. As the deformation rate increases to 0.1 s⁻¹, the high strain value regions of the fcc phase and the bcc phase increase correspondingly. When the deformation rate increases to 1 s⁻¹, the high strain value regions of the fcc phase and bcc phase show no significant increases. When the deformation rate increases to 10 s⁻¹, the high strain value regions of the fcc phase and bcc phase increase correspondingly, and the high strain grains are mainly concentrated in the fcc phase. From figure 9 as a whole, at low temperature, as the strain rate increases, the GOS value also increases, showing obvious non-uniform grain deformation. Among them, it can also be observed deformation serious bcc area of the red and white of the fcc area. This is due to the compression at 850 °C, low deformation temperature, a large number of distortion defects caused by dislocation slip, and poor diffusion ability of atoms, which can not be eliminated by DRV and DRX in time.

Figure 10 shows the GOS of two phases at different strain rates at 1000 °C, and the figure 10 has the same scale as figure 9. When the deformation temperature increases to 1000 °C, the GOS value is slightly lower than that of 850 °C (figure 9), and the grain deformation is also uneven, which may be caused by the different degree of nucleation and recrystallization of fcc phase and bcc. At this temperature, dislocation slip gathers from the soft bcc phase to the interface of the two phases, which increases the degree of distortion at the interface of the two phases and promotes the DRX nucleation of the harder fcc at the interface. With the increase of the deformation rate to 0.1 s⁻¹, the high strain regions of fcc and bcc phases increase correspondingly. When the deformation rate increases to 1 s⁻¹, the high strain regions of fcc and bcc phases do not obviously increase, and
the phases are flattened. When the deformation rate increases to 10 s$^{-1}$, the high strain regions of fcc and bcc phases increase obviously, and the high strain grains are mainly concentrated in the fcc phase.

Figure 11 shows the grain orientation spread (GOS) at different strain rates at 1050 °C. The scale in figure 11 is the same as that in figure 9. It can be seen from the GOS values in fcc and bcc phases are both low at 0.01 strain rate. With the increase of the deformation rate to 10 s$^{-1}$, the GOS values in the fcc phase and the bcc phase increase. However, the GOS values of fcc phase increased more, and its deformation degree is greater. From figures 9–11, the GOS value gradually decreases, because as the temperature increases, the atomic diffusion ability and grain boundary migration ability increase, reducing the driving energy for DRV and DRX. As a result,
DRV and DRX are generated to eliminate distortion defects caused by dislocation slip, and the GOS value is reduced. It can also be clearly seen in figure 11 that the sample deformed at 1050 °C has the smallest strain gradient.

In the EBSD test, the blue region indicates the soft regions of recrystallization and dislocation with low dislocation density. The yellow region indicates a sub-grain region with low dislocation density. The red region indicates a highly deformed region with high dislocation density. These three differences are directional difference angles (1–7.5°).

Figure 12 shows the morphological characteristics of different recrystallized regions at different strain rates at 850 °C. Figure 12 shows that as the strain rate increases, the recrystallization fraction of the bcc phase continuously decreases from 48% to 21.9%, and the deformation region of the bcc phase increases from 1.1% to 11.8%. With the increase of the strain rate, the recrystallization fraction of the fcc phase decreases slightly from 1.1% to 0.6% and remained unchanged afterwards, while the deformation region of the fcc phase increases from 36.8% to 64.2%, and the deformation rate suddenly increases at 10 s⁻¹. Since the recrystallization nucleation is mainly controlled by temperature and time, at low temperatures, low strain rate dislocations have more time to move, rearrange, nucleate, and recrystallize, and as the strain rate increases, the grain boundary migration time decreases, so the recrystallization fraction decreases and the deformation area increases.

Figure 13 shows the morphological characteristics of different recrystallized regions at different strain rates at 1000 °C. It can be seen from figure 13 that as the strain rate increases, the recrystallization fraction of the bcc phase first decreases and then increases, the recrystallization fraction decreases from 59.4% to 42.7%, and then increases to 69.9%, and the deformation region of the bcc phase increases from 0.5% to 1.1%, and remains
basically unchanged later. As the strain rate increases, the recrystallization fraction of the fcc phase decreases first from 12.2% to 2.9%, then increases slightly to 5.4%. The deformation region of the fcc phase continues to increase significantly from 11.7% to 75.9%, the subcrystal region of the fcc phase continues to decrease gradually, from 75.9% to 18.7%, and the deformation rate decreases significantly at 10 s\(^{-1}\). Although the deformation region of the fcc phase is increasing largely, from 11.7% to 75.9%, the substructure region of the fcc phase continues to decrease gradually, from 75.9% to 18.7%, and the deformation rate decreases significantly at 10 s\(^{-1}\).

Figure 14 shows the morphological characteristics of different recrystallized regions at different strain rates at 1050 °C. It can be seen from figure 14 that with the increase of the strain rate, the deformation region of the bcc phase decreases slightly from 7% to 0.9%, while the substructure fraction of the bcc phase recrystallization decreases sharply from 91.9% to 31.4%. The bcc phase recrystallization fraction increases continuously and significantly from 1.1% to 67.7%. As the strain rate increases, the recrystallization fraction of the fcc phase increases slightly from 5.9% to 17.7% while the deformation region of the fcc phase increases significantly, from 2.5% to 48.1%, and the fcc phase substructure region decreases obviously, from 91.5% to 34.3%.

It is not possible to determine which of the two phases of 2205 duplex stainless steel has recrystallized from the curve in figure 2, because the content of both phases in the experimental steel is not low, and any phase has to undergo DRX. It is generally believed that the bcc phase (ferrite phase) of the body-centered cubic structure has the higher stack fault energy, and the dynamic recovery is the main softening mechanism; while the fcc (austenite phase) of that has lower stack fault energy, and its softening mechanism is mainly DRX\(^{[19]}\). During the deformation process, the DRV softening the ferrite is prone to precede the DRX softening the austenite is prone to\(^{[20,21]}\).

Ferrite coexists with austenite in duplex stainless steels at a high temperature. Although the ferrite phase has higher interlayer fault energy, compared to the austenite phase, activities such as cross slip, climbing, etc are easy to occur, and the DRV rate is high. However, ferrite is relatively soft while austenite is relatively hard at high temperature. The deformation resistance of the ferrite phase is smaller than that of the austenite phase, so the deformation is mainly concentrated in the ferrite phase. The ferrite phase deformed energy storage increases in a thermodynamically unstable state, and the austenite hard phase distributed on the ferrite phase matrix will increase the dislocation proliferation and the deformation energy storage. As a result, the DRV, the softening mechanism of the ferrite phase, is insufficient to reduce the deformation energy storage so that the ferrite phase undergoes DRX as the softening mechanism. It can be seen from figures 5–7 that austenite is in a flat state, while ferrite undergoes DRX, grain equiaxation, and there are some recrystallized small grains. In figures 12–14, recrystallization mainly occurs in the bcc phase (ferrite phase). The DRV and DRX shown by the substructure fraction and recrystallization fraction in the bcc phase are relatively higher than fcc, while the red deformation area is almost all in fcc austenite phase. At the same time in the GOS diagram (figures 9–11), under different deformation conditions, the high strain grains of fcc are always higher than bcc. Therefore, the softening phenomenon exhibited by the reduction of the flow stress after the peak is mainly related to the ferrite bcc.
In the statistics of figures 15 (a) and (b), the recrystallization proportion of bcc phase is much larger than that of fcc phase. In addition, at a low strain rate, the bcc phase is around 940 °C with a higher degree of recrystallization, while the fcc phase is around 990 °C with a higher degree of recrystallization, because as the temperature increases, the atomic diffusion ability and the grain boundary migration ability are enhanced, and the bcc phase with high stacking fault energy is more likely to DRV while the fcc phase with low stacking fault energy is subject to cross slip, climbing, etc, making the difficulty of the activity process greater. Therefore, the low dynamic recovery rate and the large dislocation density of the substructure promote the nucleation of DRX. At high strain rates, the temperatures at which the bcc phase and the fcc phase have a large proportion of DRX are 1000 °C and 1050 °C, respectively, which have increased compared with those at the low strain rate, because at the high strain rate, the incubation time of recrystallization is short, and higher temperatures are needed to increase the atomic diffusion ability and the grain boundary migration ability. In summary, during the 2205 thermal deformation process, the bcc phase has a higher proportion of DRX than the fcc phase, and the maximum recrystallization temperature of the bcc phase is about 50 °C lower than that of the fcc phase.
4. Conclusion

In this paper, the hot deformation behavior of 2205 duplex stainless steel was studied through an equiaxed compression experiment.

(1) Through dynamic analysis, the thermal deformation activation energy \( Q = 351.58 \text{ kJ mol}^{-1} \) and the stress index \( n = 3.8543 \) were calculated. The established thermal deformation constitutive equation is:

\[
\dot{\varepsilon} = 1.18 \times 10^{10} \left[ \sinh(0.00645\sigma) \right]^{3.8543} \exp(-351580/RT)
\]

(2) At the same temperature, as the strain rate increases, the austenite phase content decreases slowly, while the ferrite phase content increases. At 850 °C with a strain rate of 0.01 s\(^{-1}\), the proportions of the bcc phase and the fcc phase are 69% and 28%, respectively. At 1050 °C with a strain rate of 10 s\(^{-1}\), the proportions of bcc and fcc are 65% and 35%, respectively. With the increase of the strain rate, the transformation from austenite to ferrite is promoted.

(3) During the 2205 thermal deformation process, the bcc phase has a higher proportion of dynamic recrystallization than the fcc phase, and the maximum recrystallization temperature of the bcc phase is about 50 °C lower than that of the fcc phase.

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