Strain Rate Sensitivity Behaviour of a Chrome-Nickel Austenitic-Ferritic Stainless Steel and its Constitutive Modelling

Amit KUMAR,1) Aman GUPTA,1) Rajesh Kisni KHATIRKAR,1)* Nitish BIBHANSHU2) and Satyam SUWAS2)

1) Department of Metallurgical and Materials Engineering, Visvesvaraya National Institute of Technology (VNIT), South Ambazari Road, Nagpur – 440010, Maharashtra, India.
2) Department of Materials Engineering, Indian Institute of Science (IISc), Bengaluru – 560012, Karnataka, India.

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In the present investigation, the plastic flow curves and work softening behaviour of a dual phase Fe–Cr–Ni alloy during hot deformation (low to intermediate temperature range, 948 K (675°C) to 1 248 K (975°C)) along with concurrent microstructural development were investigated. The flow stress increased with the increase in strain rate and decreased with the increase in deformation temperature. The single peak characteristic appearing in all the flow curves indicated that dynamic recrystallization (DRX) was the dominant softening mechanism in the later stage of deformation. The critical strain for DRX initiation was εc = 0.632εp and the peak strain (εp) were expressed through the Zener-Hollomon parameter (Z). For flow stress modelling, an Arrhenius type constitutive model was established to predict the flow stress behaviour during hot deformation. The results showed that the calculated flow curves agreed reasonably well with the experimental results. The microstructural analysis using optical microscopy indicated that all the deformed structures exhibited elongated grains similar to that of parent microstructure and some equiaxed grains (resulting from DRX in the austenite phase). The fraction of equiaxed grains (in austenite) increased with the deformation temperature. At low Z, the ferrite phase accommodates the strain and dynamic recovery was the prominent restoration process. At high Z, austenite controlled the deformation mechanism and DRX was the likely cause for microstructural refinement. The iso-strain rate sensitivity (m) contour map was used to determine the optimum regime of high temperature workability.

KEY WORDS: Fe–Cr–Ni alloy; dynamic recrystallization; dynamic recovery; softening mechanism; critical strain; Zener-Hollomon parameter; Arrhenius type constitutive model.

1. Introduction

The alloys based on Fe–Cr–Ni system have a microstructure consisting of austenite (FCC) and ferrite (BCC) and are characterized by good mechanical properties, stress corrosion cracking resistance and weldability.1,2) The proportion of austenite and ferrite in the duplex microstructure is dependent on the alloying additions and thermo-mechanical processing conditions.1,3–5) The alloys based on the Fe–Cr–Ni system (UNS S32205, UNS S32750, UNS S32760, UNS S32304 and UNS S32101) undergo a series of processing operations like casting, hot rolling cold rolling, solution annealing, hot compression, hot extrusion, hot forging before being put into the final application.4,6) Due to their very high strength at room temperature, these alloys are processed at high temperatures (above approximately 0.6T_m, T_m is the melting point).7) The hot working is a critical step as it should not lead to defects (like cracks, uneven surface finish) which can deteriorate the mechanical properties and also should develop a microstructure for satisfactory product performance (in terms of mechanical properties and corrosion resistance).8) The mechanical properties of these steels are dependent on the austenite-ferrite phase proportion, crystallographic texture and presence or absence of intermetallics (σ, χ, Lave phases, nitrides and carbides).9) During deformation at temperatures above ambient, both static and dynamic recovery, static and dynamic recrystallisation can take place depending on the thermo-mechanical processing variables.10) The deformation and softening behaviour of ferrite and austenite in these alloys depends on their stacking fault energy (SFE). The SFE has been reported11) to be a function of crystal structure and chemical compositions. Therefore, by changing the chemical composition, the properties of these alloys can be tailored (e.g. the transformation induced plasticity (TRIP) effect12)). Similarly, the softening mechanisms can also get modified and hence each alloy in Fe–Cr–Ni system has a different behaviour. It is well reported13,14) that due to high SFE of δ-ferrite, it undergoes dynamic recovery (DRV) and has a better high temperature workability than austenite. Austenite, on the other hand, has low SFE and hence undergoes DRV only to limited extent. When the dislocation density reaches a critical value corresponding to a critical strain (εc), the initiation of DRX takes place15) resulting in flow softening. Fe–Cr–Ni alloys have, in general, poor hot workability which can be attributed to the different deformation/softening response of each constituent phase and their interaction in the duplex microstructure. In the hot working temperature range, austenite is significantly stronger than ferrite, which affects the load/strain transfer between both the phases and hence their hot strength and ductility. The existence of austenite as a harder and δ-ferrite...
as sofer phase at elevated temperatures is likely to results in strain partitioning at the early stage of hot deformation. At higher strains, load is transferred from ferrite to austenite leading to increased dislocation density and work hardening (WH) until DRX is initiated.\textsuperscript{15} The hot deformation behaviour of commercial grades of duplex stainless steels like UNS S32205, UNS S32507 has been studied.\textsuperscript{8,13,15-17} Liu et al.\textsuperscript{18} suggested that continuous dynamic recrystallisation (CDRX) was the restoration mechanism for both ferrite and austenite phases in UNS S32101 duplex stainless steel (DSS). Fan et al.\textsuperscript{19} reported that during the hot deformation of cast UNS S32205 DSS, ferrite was softened by DRX, while austenite softened by DRV without DRX. Ma et al.\textsuperscript{20} suggested that during hot deformation of UNS S32205 DSS, the softening mechanism in each constituent phase was different. Ferrite gets softened by DRV and CDRX, while the austenite undergoes discontinuous dynamic recrystallisation (DDRX) to a limited extent. The investigations on hot deformation behaviour of materials are primarily based on two approaches, (i) the physical modelling approach and (ii) the phenomenological/empirical model approach.\textsuperscript{6,21} The first approach is related to the use of various physical parameters like grain size and dislocation density to describe the phenomenon. Such an approach is more reliable in describing microstructural evolution and various interconnected metallurgical phenomenon such as work/strain hardening (WH/SH), DRV and DRX during hot deformation. However, these models require extensive and costly experiments to identify material parameters.\textsuperscript{22} The second approach is phenomenological, which provides a mathematical correlation between flow stresses and various deformation parameters under a wide range of working conditions. The material parameters for such models can be determined from the experimental stress-strain curves. These models are simple yet reliable in predicting the hot deformation behaviour of materials for engineering applications.\textsuperscript{9} Farnoush et al.\textsuperscript{13} and Momeni et al.\textsuperscript{23} studied the hot deformation behaviour of UNS S32205 DSS and showed the effect of deformation temperature and strain rate by using Zener–Hollomon parameter. They found that the material constants differ significantly at low and high temperatures. Kingklang et al.\textsuperscript{24} studied the hot deformation of UNS S32507 DSS and showed a strong dependency of high temperature deformation behaviour on processing parameters. Li et al.\textsuperscript{25} investigated the hot deformation behaviour of UNS S32707 hyper DSS and suggested that flow stress curves at temperatures higher than 938 °C and strain rate less than 1 s\textsuperscript{-1} which the DRV in ferrite as the softening mechanism without peak stress, while at lower temperatures DRX remains the dominant softening mechanism in austenite. In most of these studies,\textsuperscript{13,16,23} the temperatures used were above 1 073 K (800°C). The aim of the present work is to study the stress-strain behaviour and the corresponding microstructure evolution using compression tests in the temperature range 948 K (675°C) to 1 248 K (975°C). The flow stress modelling, Arhenius hyperbolic sine law\textsuperscript{26} was used.

2. Experimental

The chemical composition of the material used in the present investigation was Fe-0.02%C- 22.2%Cr- 5.3%Ni-3.3%Mo. It was received in the form of bar of diameter 150 mm in the hot rolled condition. Prior to hot compression tests, the as-received Fe–Cr–Ni alloy was solution annealed (SA) at 1 348 K (1 075°C) for 4 hrs in an inert atmosphere furnace to homogenize the chemical composition. Small cylindrical samples of 10±0.02 mm diameter and 15±0.05 mm height were cut with the axis parallel to the original rolling direction (RD) using electronic discharge machining (EDM). The end surfaces of specimens were made flat and parallel. The hot compression behaviour was investigated using isothermal compression test performed on 100 kN servo-hydraulic test machine (DARTEC, UK). The cylindrical samples were placed in the machine and heated to the test temperature and held there for 20 minutes to attain the uniform temperature throughout the sample. All the samples were compressed along the axis to 50% height reduction (corresponds to a true strain of −0.69). Four different temperatures (948 K (675°C), 1 048 K (775°C), 1 148 K (875°C) and 1 248 K (975°C)) and three strain rates (0.01, 0.1 and 1 s\textsuperscript{-1}) were used constituting a total of 12 deformation conditions. Graphite powder was used as lubricant on both the ends of the sample to minimise the friction to ensure effective and uniform deformation. After uniaxial compression, the deformed samples were immediately quenched in water to preserve the microstructure. For metallographic investigation, the deformed specimens were sliced parallel to the compression axis. For optical microscopy (OM), all the samples were polished using the standard metallographic procedure followed by etching with Kalling’s reagent (a solution of 5 gm CuCl\textsubscript{2}, 100 ml hydrochloric acid and 100 ml ethanol).\textsuperscript{20} Micro-hardness was measured for both the phases (ferrite and austenite) using Vickers micro-hardness tester at load of 300 gm. The microstructure of the initial SA sample consisted of two phases, ferrite matrix in which elongated austenite grains were oriented along the original hot RD (as shown in Fig. 1(a)).

3. Results and Discussion

3.1. Flow Characteristics

The true stress-strain curves of Fe–Cr–Ni alloy used in the present investigation at different deformation conditions is shown in Figs. 1(a) to 1(d). The flow stress is strongly dependent on the processing parameters viz., deformation temperature, strain rate and strain. The flow stress increased with the increase in strain rate or decrease in temperature as shown in Figs. 1(e) and 1(f) respectively. The increased peak stress with strain rate indicates that more dislocation are formed which promote work hardening at higher strain rates.\textsuperscript{27} The effect of deformation temperature on the change in flow stresses was more pronounced than that of strain rate. These results corroborate the results of Kingklang et al.\textsuperscript{24} and Farabi et al.\textsuperscript{25} In order to find the effect of strain rate sensitivity index (m) and strain hardening exponent (h) on the hot deformation behaviour, following equation was used:

\[
\sigma = C e^{k} e^{m} \tag{1}
\]

where C is a constant. Non-linear square regression was used to determine the values of m and h at different temperatures and these values are given in Table 1. The value of h decreased while m increased with increase in temperature. This could be attributed to the changes in SFE of the material with temperature - at high temperatures SFE increases making the dislocation movement easy reducing the strain hardening of the material during deformation. In early stage of hot deformation, flow stresses rapidly increased due to strain/work hardening at all deformation temperatures and strain rates. Work hardening was perceived to have taken place due to dislocation pileup and was more significant at lower temperatures (948 K or 675°C) and higher strain rates (1 s\textsuperscript{-1}). The work hardening rate gradually decreased until the flow stress reached its peak value (\(\sigma_{p}\)). Beyond this peak value, the flow stress decreased with further increase in strain, indicating the occurrence of softening mechanism(s) (DRV and DRX) depending upon the deformation parameters mainly the temperature. The presence of \(\delta\) ferrite with high SFE and 48 slip systems in the vicinity of austenite with low SFE and 12 slip systems is expected to affect the hot deformation behaviour significantly. Ferrite was prone
to DRV,\(^{16}\) while DRX was the dominant restoration mechanism in austenite.\(^{16}\) At room temperature, ferrite was harder than austenite and hence during deformation at room temperature; austenite takes up the strain first. The substructure development/shear bands was reported to be observed in the austenite grains even at very low cold rolling percent.\(^{28,29}\) In contrast, at high temperatures, the ferrite was softer than the austenite.\(^{15}\) Initially at low deformation levels, the strain was mostly accommodated by ferrite. At higher deformations, the load was likely to be transferred from ferrite (softer phase) to austenite (harder phase) leading to strain accumulation in austenite till the initiation of DRX. It was reported\(^{15,20}\) that for low temperatures (below 1 373 K or 1 100°C) and at intermediate strain rates, DRV in ferrite was restricted and rapid load transfer facilitates the onset of DRX in austenite. In the present investigation, the deformation was carried out below 1 373 K (1 100°C). Further, all the stress-strain curves showed similar characteristics with a single peak stress, which has also been classically attributed to onset of DRX as the dominant restoration mechanism.\(^{30}\) Once the dynamic equilibrium between the work hardening and softening occurs, a steady state stress (\(\sigma_{\text{ss}}\)) is achieved.

\[Z = \dot{\varepsilon} \exp \left( \frac{Q}{RT} \right) \]  

(2)

3.2. Constitutive Equations

Different constitutive equations have been proposed in published literature to model the deformation behaviour of various materials at high temperatures.\(^{8,13,23,32,33}\) The hyperbolic sine Arrhenius equation has been widely used to describe relationship between flow stress and processing parameters.\(^{8,25,34-38}\) Irrespective of the restoration mechanism(s) involved, the effect of processing parameters on the deformation behaviour of material can be expressed by Zener-Hollomon parameter (Z):

\[Z = \dot{\varepsilon} \exp \left( \frac{Q}{RT} \right) \]  

(2)
The strain rate ($\dot{\varepsilon}$) and temperature (T) dependence on stress ($\sigma$) is expressed as:

$$\dot{\varepsilon} = AF(\sigma)e^{-\frac{Q}{RT}}$$ \hspace{2cm} (3)

where $F(\sigma)$ is the stress function, which could be expressed by one of the following:

$$F(\sigma) = \begin{cases} \sigma^n & : \sigma < 0.8 \\ \exp(\beta\sigma) & : \sigma < 1.2 \\ \sinh(\alpha\sigma)^n & : \text{for all} \end{cases}$$ \hspace{2cm} (4)

Substituting Eq. (4) into (3), the following equations were obtained:

$$\dot{\varepsilon} = A\sigma^n \exp\left(-\frac{Q}{RT}\right)$$ \hspace{2cm} (5)

$$\dot{\varepsilon} = A\exp(\beta\sigma)\exp\left(-\frac{Q}{RT}\right)$$ \hspace{2cm} (6)

$$\dot{\varepsilon} = A[n\sinh(\alpha\sigma)]^n \exp\left(-\frac{Q}{RT}\right)$$ \hspace{2cm} (7)

where, $Q$ (J/mol) and $R$ (8.314 J/mol K) are the hot working activation energy and universal gas constant respectively, T (K) is the absolute temperature, $\dot{\varepsilon}$ (1 s$^{-1}$) is the strain rate, $\sigma$ (MPa) is the flow stress. A, $Q$, $\beta$, $\alpha$ and $n$ are material constants in which $\alpha = \beta/n$. For wide range of stresses, the hyperbolic sine function is more acceptable and hence used in the present investigation. Initially, the peak stresses were taken for each combination of temperature and strain rate to determine the material constants. For modeling of the constitutive equations, four material constants ($\alpha$, $n$, $Q$ and $A$) were determined from experimental flow curves as follows:

a) Determination of $\alpha$.

The value of $\alpha = \beta/n'$, the value of $\beta$ and $n'$ could be determined by taking natural log on both sides of Eqs. (5) and (6). For all the calculations, stress taken was the peak stress ($\sigma = \sigma_p$):

$$\ln\dot{\varepsilon} = n'\ln\sigma_p + \ln A - \frac{Q}{RT} \hspace{2cm} (8)$$

$$\ln\dot{\varepsilon} = \beta\sigma_p + \ln A - \frac{Q}{RT} \hspace{2cm} (9)$$

At constant temperature the partial difference of Eqs. (8) and (9) will be expressed as:

$$\left[\frac{d\ln\sigma_p}{d\ln\dot{\varepsilon}}\right]_{T} = \frac{1}{n'} \hspace{2cm} (10)$$

$$\left[\frac{d\sigma_p}{d\ln\dot{\varepsilon}}\right]_{T} = \frac{1}{\beta} \hspace{2cm} (11)$$

By linear regression, the average slope of $\ln\sigma_p$ versus $\ln\dot{\varepsilon}$ and $\sigma_p$ versus $\ln\dot{\varepsilon}$ will give the value 1/$n'$ and 1/$\beta$ as shown in the Figs. 2(a) and 2(b) respectively. The average value of $n'$ is 12.74 and $\beta$ is 0.057. Therefore, the value of $\alpha$ was approximately 0.0045.

b) Determination of $n$:

To calculating the value of $n$, natural log of Eq. (7) was taken:

$$\ln\dot{\varepsilon} = n'\ln\sinh(\alpha\sigma_p) + \ln A - \frac{Q}{RT} \hspace{2cm} (12)$$

At constant temperature during hot working, the partial differentiation of Eq. (12) will give the following:

$$\left[\frac{d\ln\sinh(\alpha\sigma_p)}{d\ln\dot{\varepsilon}}\right]_{T} = \frac{1}{n} \hspace{2cm} (13)$$

The slope of $\ln[\sinh(\alpha\sigma_p)]$ versus $\ln\dot{\varepsilon}$ at different temperatures will give the value of 1/n as shown in Fig. 2(c). Therefore, the average value of $n_1$ was obtained as 9.33.

c) Determination of activation energy ($Q$):

The activation energy can be calculated by rearranging the Eq. (12) as follows:

$$\ln\sinh(\alpha\sigma_p) = \frac{Q}{n_1RT} + \frac{n_1}{n_1} - \ln A \hspace{2cm} (14)$$

Now, at constant strain rate, the partial differentiation of Eq. (14) at different temperatures will give the following equation:

$$\left[\frac{d\ln[\sinh(\alpha\sigma_p)]}{d(1/RT)}\right]_{\dot{\varepsilon}} = \frac{Q}{n_1R} \hspace{2cm} (15)$$

Thus the activation energy can be calculated as:

$$Q = n_1R\left[\frac{d\ln[\sinh(\alpha\sigma_p)]}{d(1/RT)}\right]_{\dot{\varepsilon}} \hspace{2cm} (16)$$

The average slope of $\ln[\sinh(\alpha\sigma_p)]$ versus (1 000/7) at different strain rate is 7.25 (refer Fig. 2(d)). Using Eq. (16), the value of $Q$ was ~561.774 kJ mol$^{-1}$. The value of activation energy ($Q$) approximately describes the sum of all the energies needed to overcome the activation barrier during hot deformation. The energy barriers typically refers to one or more of the processes like climb of edge dislocations, overcoming of the forest of dislocations, cross slip, movement of dislocations, lattice distortion caused by solute atoms etc.$^{5(3)}$

The value of $Q$ decreased with increase in the deformation temperature. This could be attributed to higher diffusion rates and less energy requirement for crossing the activation barrier. The activation energy ($Q$) values reported in the literature for materials with similar composition are in the range 432–460 kJ mol$^{-1}$. However, the average value obtained in present investigation (i.e. 561.774 kJ mol$^{-1}$) differs due to wider temperature range (948–1 248 K/675–975°C), while most of the values reported in the literature were for high temperature deformation (1 173–1 473 K or 900–1 200°C). It is anticipated that, in the present investigation, the Q value for the deformation of dual phase steel should be approximately the average of activation energies for hot deformation of BCC and FCC single phase materials and also significantly lower than the value for FCC.$^{5(3)}$

However this is not always the case, for materials like 2 507 duplex stainless steel, the activation energy values were reported to be significantly higher than the FCC materials.$^{5(3)}$ In highly alloyed steels, the value of $Q$ depends on the alloying content, precipitate size distribution and phase transformation during deformation. All these factors affect the dislocation activity and hence the active deformation mechanism.$^{5(3)}$

d) Determination of $A$:

The value of $A$ could be estimated by substituting Eq. (7) into (2):

$$Z = A[n\sinh(\alpha\sigma_p)^n] \hspace{2cm} (17)$$

The Zener-Hollomon ($Z$) parameter was estimated from Eq. (1) for all temperatures and strain rates. $Z$ increased with decreasing temperature and increasing strain rate. Taking natural log of both sides of Eq. (17), we get:

$$\ln Z = n\ln[\sinh(\alpha\sigma_p)] + nA \hspace{2cm} (18)$$
The relationship between $\ln Z$ versus $\ln [\sinh(\beta \sigma_p)]$ is plotted in Fig. 2(e). The slope and intercept of this plot gives the value of stress exponent ($n$) and $\ln A$ respectively. Thus, the value of $n$ was $\approx 9.244$ and $A = 2.38 \times 10^{24}$.

Table 2 shows the obtained values of all the material constants at peak stress. The hyperbolic sine function for the present Fe–Cr–Ni alloy was obtained as:

$$Z = \dot{\varepsilon} \exp\left(\frac{561.774}{8.314\, T}\right) = 2.38 \times 10^{24} [\sinh(0.0045\sigma_p)]^{9.33} \ldots \quad (19)$$

The values of all the material constants ($\alpha$, $n$, $Q$, $A$) were calculated based on the above equations following a similar procedure for different strains with an interval of 0.05 in the range 0.05–0.7. The calculated material constants are plotted with increasing strain in Fig. 3. It could be seen that the polynomial function of 6th order shows a good relationship between these constants and strain. The 6th order polynomial equations are as follows:

$$\alpha = a_0 + a_1\dot{\varepsilon} + a_2\dot{\varepsilon}^2 + a_3\dot{\varepsilon}^3 + a_4\dot{\varepsilon}^4 + a_5\dot{\varepsilon}^5 + a_6\dot{\varepsilon}^6 \ldots \quad (20)$$

$$n = b_0 + b_1\dot{\varepsilon} + b_2\dot{\varepsilon}^2 + b_3\dot{\varepsilon}^3 + b_4\dot{\varepsilon}^4 + b_5\dot{\varepsilon}^5 + b_6\dot{\varepsilon}^6 \ldots \quad (21)$$

$$Q = c_0 + c_1\dot{\varepsilon} + c_2\dot{\varepsilon}^2 + c_3\dot{\varepsilon}^3 + c_4\dot{\varepsilon}^4 + c_5\dot{\varepsilon}^5 + c_6\dot{\varepsilon}^6 \ldots \quad (22)$$

$$\ln A = d_0 + d_1\dot{\varepsilon} + d_2\dot{\varepsilon}^2 + d_3\dot{\varepsilon}^3 + d_4\dot{\varepsilon}^4 + d_5\dot{\varepsilon}^5 + d_6\dot{\varepsilon}^6 \ldots \quad (23)$$

Now, Eq. (17) can be expressed in the terms of flow stress as:

![Fig. 2.](image-url)
\[ \sigma = \frac{1}{a} \sinh^{-1} \left[ \frac{Z}{A} \right] \]  \hspace{1cm} (24)

and the sin inverse function can be expressed as:

\[ \text{Arcsinh} (x) = \ln \left[ x + (x^2 + 1)^{1/2} \right] \]  \hspace{1cm} (25)

Substituting \( x = \left( \frac{Z}{A} \right)^{1/n} \) in Eq. (25), plastic flow stress can be expressed as the hyperbolic sine function in terms of Zener–Hollomon parameters (Z) to predict the behaviour of the investigated steel at different true strains (\( \varepsilon_i \)).

The flow behaviour of metals and alloys is largely dependent on the initial microstructure and microstructural changes that take place during deformation. The optical microstructures (bright field) of samples deformed to 0.6 true strain at different temperatures and strain rates are shown in Fig. 5 (austenite = light colour and \( \delta \)-ferrite = dark colour). After deformation, no macroscopic defects (cracks) were observed in the samples and grain size of austenite in the microstructure became more refine and equiaxed with increase in deformation temperature. The phase fractions did not change significantly with the change in deformation conditions. At low deformation temperature (948 K or 675°C), all the grains exhibited elongated morphology. At high temperatures \( i.e. 1148 \text{ K (875°C) and 1248 K (975°C)} \), the grain possibly recrystallized and an appreciable amount of deformed grains were retained. A more detailed investigation based on electron backscattered diffraction (EBSD) would be needed to comment on it. Figure 6 shows the changes in micro-hardness of austenite and ferrite as a function of deformation condition. The hardness values at different conditions were used to approximately characterise the softening mechanism \( i.e. \) DRV and DRX.

It has been reported that hardness decreases with increase in the deformation temperature. In the present investigation, the decrease in ferrite hardness was minimal. It could...
therefore, be speculated that ferrite was less dominant during deformation in the present Fe–Cr–Ni alloy. This indicates a reduced role of DRV as a softening behaviour under investigated processing parameters. On the other hand, the hardness of austenite showed significant difference and decreased for all strain rates as the deformation temperature increased. The relative hardness of ferrite and austenite showed a decreasing trend for all the conditions with increase in temperature. This is likely to be due to the dominance of DRX mechanism in the austenite as restoration process. In dual phase Fe–Cr–Ni alloys, ferrite has high SFE and austenite has low SFE. Due to difference in the SFE, ferrite and austenite have different responses to the hot deformation conditions. Ferrite having higher SFE tends to soften by DRV. The rate of strain accumulation in the microstructure depend on strain rate, temperature and SFE. In high SFE materials, the dislocation movement is easier due to easy cross slip and dislocation climb. In low SFE materials, cross slip is restricted and recrystallization mechanism is favoured during hot deformation. The SFE of austenite in present alloy was 19.1 mJ m\(^{-2}\), which is in the range of low SFE (~10 to 29 mJ m\(^{-2}\)). The lower SFE of austenite and almost 50% phase fraction is expected to promote DRX as the main restoration mechanism, if dislocation density reaches a critical value.

Fig. 4. (a) through (d) Comparison between true stress-strain curves determined experimentally and those calculated by the modelled Zener-Hollomon equation and (e) correlation between the predicted and experimental flow stresses for different deformation conditions.

Fig. 5. Optical microstructures (bright field) of Fe–Cr–Ni alloy specimen after hot compression tests at different temperatures and strain rates. C.D. shows the compression direction.
3.4. Critical Conditions for Onset of DRX

The flow stress curves give an indication that after some critical strain, DRX can occur in austenite, however, they do not provide exact information about the onset/initiation of DRX. The critical strain can be determined by microstructural observation of the deformed sample. It is most reliable, but more difficult due to phase transformation and requirement of large number of experiments for precise determination. Poliak and Jonas,40) Najafizadeh and Jonas,41) Ryan and McQueen34) studied the flow stress behaviours for deformation at constant \( \dot{\varepsilon} \) and showed that the initiation of DRX can be attributed to the inflection point in the strain hardening rate \( \theta \) (the differentiation of stress with respect to the strain \( \theta = d \sigma / d \varepsilon \)) vs flow stress \( (\sigma) \). The \( \theta-\sigma \) curves for various processing conditions are shown in Figs. 7(a) to 7(d). The work hardening curves typically showed two regions as the slope was gradually changed to a lower slope and then dropped towards \( \theta = 0 \) indicating the peak stress \( (\sigma_p) \). The inflection point on the \( \theta-\sigma \) curves was attributed to the onset of DRX. The first order differentiation of \( \theta-\sigma \) (i.e. \( \delta \theta / \delta \sigma \sigma \)) curves has a unique maximum value at the inflection point of the work hardening curves. Mathematically, a null value of the second order differentiation of \( \theta \) with respect to \( \sigma \) will indicate the inflection point and the corresponding stress and strain will be called as critical stress \( (\sigma_c) \) and critical strain \( (\varepsilon_c) \) respectively (Figs. 7(a) to 7(d)).

\[
\frac{\delta^2 \theta}{\delta \sigma^2} = 0 \quad \text{.................................. (29)}
\]

The ratios \( \sigma_c/\sigma_p \) and \( \varepsilon_c/\varepsilon_p \) are known as the critical ratios for onset of DRX. The critical condition \( (\sigma_c, \sigma_p, \varepsilon_c, \varepsilon_p) \) versus \( Z \) are plotted on logarithmic scale in Figs. 7(e) and 7(f). The critical stress ratio \( \sigma_c/\sigma_p \) and critical strain ratio \( \varepsilon_c/\varepsilon_p \) are found to be 0.96 and 0.632 respectively. The predicted relation of peak stress and strain with \( Z \) can be expressed using power-law function as per Eq. (30):

\[
\varepsilon = xZ^y \quad \text{.................................. (30)}
\]

Where \( x \) and \( y \) are constants. Now the developed equations will be:

\[
\sigma_p = 4.65Z^{0.0672} \quad \text{........................... (31)}
\]

\[
\varepsilon_p = 0.165Z^{0.011} \quad \text{........................... (32)}
\]

The critical strain ratio \( (\varepsilon_c/\varepsilon_p = 0.632) \) of present Fe–Cr–Ni alloy was found to be higher than critical strain ratio of single phase austenitic steels \( (\varepsilon_c/\varepsilon_p = 0.48–0.52) \). This is expected to be due to the presence of two phases in the microstructure as compared to the single-phase austenitic steel. The applied strain gets partitioned into both the phases and both have different crystal structure and deformation response.23) The occurrence of DRX has been reported to be observed at the austenite/austenite grain boundaries.7) The frequency of these boundaries is less in the present Fe–Cr–Ni alloy due to the banded austenite-ferrite structure (Fig. 1(a)). Techovnik et al.42) found the occurrence of DRX in duplex steels at very high deformation temperature of 1 523 K which is beyond the temperature range used in the present investigation and hence is not expected.

3.5. Optimum Hot Workability

The optimum processing conditions (temperature, strain rate and strain) was determined using a material parameter \( m \) (i.e. strain rate sensitivity (m). The ‘m’ parameter is rate of change of stress with strain rate at constant levels of strain \( (\varepsilon) \) and temperature \((T)\).

\[
m = \frac{\partial (\log \sigma_p)}{\partial (\log \dot{\varepsilon})} \quad \text{........................... (33)}
\]

The m value should not be negative or null, which would lead to flow instability and fracture.8,43) Most of the materials qualify this criterion during hot working condition. Higher m values ensures strain rate hardening as the tendency of localised deformation decreases, which enable extensive elongation of the material without necking during tension46,47) and prevents instabilities during compression.48)
\( m = 1 \) for metals and alloys represents ideal superplastic behaviour. For optimum workability, the region where strain rate sensitivity is high is usually chosen. For the ease, 2-D contour map in Fig. 8 has been plotted by calculating strain rate sensitivity by taking the first order derivative at each point of the \( \log \sigma_p - \log \dot{\varepsilon} \) and equal \( m \) values joined together to give iso-strain rate sensitivity contour. Distinct regions i.e. domains of low \( m \) and regions of high \( m \) can be marked in these contour maps (Fig. 8) indicating that the deformation mechanism varies with processing parameters. This variation was more prone to temperature change as compared to strain rate. The high strain rate sensitivity domain appeared at higher temperature region from 1148 K (875°C) – 1248 K (975°C) for entire range of strain rates (0.01, 0.1, 1 s\(^{-1}\)). In this domain, the \( m \) value varied from 0.11 to 0.16 and was considered as the optimum regime for hot workability. For the entire range of temperature and strain rate in this study, \( m > 0 \) and thus present Fe–Cr–Ni alloy was not expected to show flow instability in this deformation temperature and strain rate range.45,49)

4. Summary and Conclusions

The flow behaviour of a two phase Fe–Cr–Ni alloy was investigated by performing uniaxial compression tests in the temperature range 948–1248 K (675–975°C). The analysis of the results led to the following conclusions:

(1) All the flow curves under all testing condition showed the similar behaviour – peak stress increased with decrease in deformation temperature and increase in strain rate.
(2) The average value of material constants and deformation activation energy ($\alpha$, $\beta$, $n$, $Q$, and $A$) were determined at peak stress. A constitutive model was developed based on Arrhenius hyperbolic sine equation to predict the flow behavior under specified conditions. The average value of activation energy ($Q$) was ~561.774 kJ mol$^{-1}$.

(3) Based on the developed model, all material parameters i.e. $\alpha$, $\beta$, $n$, $Q$, and $A$ were incorporated with strain compensation with sixth order polynomial fitting. The predicted flow stress for all the deformation conditions were compared with the experimental results. The value of correlation coefficient, $R^2 = 0.994$ and average absolute relative error (AARE) of 5.432%, respectively were obtained which showed precise and good reliability of the used constitutive equation.

(4) The micro-hardness showed that dynamic recrystallization (DRX) could be the restoration mechanism in austenite. The onset of DRX was observed for all the strain rates (0.01–1 s$^{-1}$) in austenite. The deformation activation energy ($Q$) was ~561.774 kJ mol$^{-1}$. The pre-exponential factor ($A$) was calculated from the slope of the Arrhenius hyperbolic sine equation to predict the deformation temperature and strain rate. According to iso-strain rate sensitivity map, the 1 148 K (875°C) for all the strain rates was recommended as the optimum regime for hot working of the Fe–Cr–Ni alloy.

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