Hot Deformation Behavior of Fe-2%Si

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The hot deformation behavior of a high purity binary ferritic Fe-2 mass% Si alloy, undergoing no phase transformation or precipitation reactions, was investigated by hot torsion tests. The parameters for the constitutive equation of Fe-2 mass% Si steel were experimentally determined from the stress-strain curves. The activation energy for hot deformation was close to that for lattice self-diffusion in α-Fe, indicating that hot deformation of Fe-2 mass% Si steel occurs by a thermally-activated, diffusion-controlled dislocation climb mechanism. The electron backscattering diffraction technique and transmission electron microscopy were used to analyze the microstructural changes occurring during hot deformation. It was found that the hot deformed microstructure consisted of equiaxed crystallites surrounded by a high proportion of high angle boundaries. These boundaries were homogeneously distributed in the microstructure. A gradual increase in the fraction of boundaries with a misorientation in the range of 10° to 18° was observed in samples obtained from interrupted torsion tests. The formation of high angle boundaries at temperatures above 0.5Tm is attributed to a gradual increase of the misorientation of low angle boundaries by the absorption of dislocations.

KEY WORDS: hot deformation; constitutive equation; dislocation climb; continuous dynamic recrystallization.

1. Introduction

Ferritic Fe–Si steel containing more than 2 mass% Si does not have a γ→α allotropic phase transformation. This property makes it possible to study the high temperature deformation behavior, e.g., dynamic recovery and dynamic recrystallization, over a wide temperature range without undesirable microstructural changes due to the γ→α transformation upon cooling to room temperature. The hot deformation behavior of Fe–Si alloys is industrially relevant as it is related to the microstructure-controlling processes occurring during the hot rolling of electrical steel.1) It is therefore worthwhile to investigate the hot deformation behavior of ferritic model alloys, such as binary Fe-2 mass% Si steel, to improve the current understanding of the microstructural evolution during high temperature deformation.

It had been generally accepted that ferritic steel, such as Si electrical steel and ferritic stainless steels, undergo only Dynamic Recovery (DRv) during hot deformation.2) Because of their high stacking fault energy (SFE), these steels are known to be susceptible to recovery process as results of easy cross slip and dislocation climb.

On the other hand, many studies have reported experimental evidence for the occurrence of dynamic recrystallization in ferritic steel.3–7) Dynamic recrystallization in ferritic steel occurs by a mechanism which is different from the classical mechanism of Discontinuous Dynamic Recrystallization (DDRx) observed in low SFE FCC metals. Gourdét et al.8) suggested that metals with a high SFE can undergo dynamic recrystallization during high temperature deformation by the mechanism of Continuous Dynamic Recrystallization (CDRx). Because of the pronounced dynamic recovery in these metals, new grains are not generated by the standard nucleation mechanism, i.e., DDRx. Instead, low angle boundaries (LABs) between sub-grains are progressively changed into high angle boundaries (HABs) by the continuous accumulation of dislocations. This process results in the formation of new equiaxed crystallites surrounded by HABs within the deformed original grains. The CDRx microstructure is therefore characterized by aggregates of crystallites surrounded partly by LABs and partly by HABs. The development of this particular microstructure was confirmed by means of Electron BackScattering Diffraction (EBSD) in Ti-stabilized Interstitial-Free (IF) steel by Oudin et al.9)

The operation of the CDRx mechanism in high SFE metals, however, has been questioned by some researchers.10) Instead, it was proposed that the formation of equiaxed grains in high SFE metals was due to the development of serrations at original grain boundaries formed by the mechanism of Geometric Dynamic Recrystallization (GDRx).10–13)

GDRx is a process in which, in the initial stages of hot deformation, the original grain boundary becomes serrated due to the preferential formation of sub-grains at the grain boundary. With increasing strain, the original grain structure becomes elongated and the grain boundary spacing is reduced to about three times the sub-grain size. This results in the pinch-off of the original grain boundaries. As a result, equiaxed crystallites are formed of which one-third of the...
grain boundaries are surrounded by HABs. These HABs are the original grain boundaries in the initial microstructure. The misorientation of LABs, which make up two-thirds of the grain boundary, does not exceed 8°. Evidence for the operation of the GDRx mechanism on the formation of equaxed crystallites has been found in Al, a high SFE FCC metal. On the other hand, Gourdet et al. has shown that the original grain size could not be equal to three times the subgrain size after a strain of 1.5, and he has questioned the fact that the formation of the HABs observed in Al could be attributed to the GDRx mechanism.

It is, therefore, still unclear whether the HABs in a high SFE ferritic steel are formed by the gradual increase of the misorientation of LABs or whether the HABs are simply a result of a change in the geometry of the original grain boundaries. One of the practical difficulties in studying the hot deformation behavior is that the deformation behavior and resulting microstructural evolution strongly depend on the experimental conditions, such as the alloy purity and the presence of precipitates. It was reported that dynamic softening of high purity Al (99.999%) was also controlled by DRx rather than DRv, but that this was not the case for lower purity Al. Furthermore, the results of studies on the hot deformation behavior of Fe–Si alloys are inconsistent. This can be attributed to the addition of other alloying elements which affect the deformation behavior at high temperature by the formation of precipitates. It is generally accepted that precipitation promotes dynamic recrystallization by retarding the recovery process during the hot deformation. It is therefore still unclear whether the dynamic recrystallization of ferritic alloys can take place without any precipitation or solution hardening effects during hot deformation.

The main focus of the study of the hot deformation behavior of metals and alloys was originally centered on FCC metals with a wide range of SFEs. The established mechanisms were then extended to explain the hot deformation behavior of BCC metals and alloys. The mechanism established for high SFE FCC metals and alloys cannot be directly applied to BCC metals and alloys without considering essential differences in their physical properties, such as the diffusivity of atoms and the operative slip systems.

In the present contribution, the hot deformation behavior of a high purity binary ferritic Fe–2 mass% Si alloy, having no phase transformation or precipitation, was therefore investigated. The constitutive equation of Fe–2 mass% Si steel was obtained by hot torsion tests. EBSD and Transmission Electron Microscopy (TEM) were used to observe the microstructural changes which took place during hot deformation.

2. Experimental Procedure

A Fe–2.2 mass% Si alloy was prepared by melting high purity alloy constituents in a vacuum induction furnace. The chemical composition of the model alloy is listed in Table 1. Elements other than Si were controlled to extremely low concentration to avoid undesirable precipitation phenomena during the high temperature deformation.

The as-cast ingot was cut in blocks with dimensions of 150 mm × 150 mm × 250 mm, and hot rolled to 20 mm in thickness at 1050°C. Torsion specimens, 10 mm in length and 8 mm in diameter, were machined from the hot rolled plate. The specimens were heated to 1100°C and maintained for 300 seconds to obtain a homogeneous microstructure prior to the hot torsion test. This process yielded a fully ferritic microstructure shown in Fig. 1 with an average grain size, measured by linear intercept method, of 730 μm.

The absence of a phase transformation in the range of deformation temperature was verified by recording the dilatation and thermal expansion coefficient during heating and cooling by dilatometry. Rod-type dilatometry specimens, 10 mm in length and 4 mm in diameter, were machined from hot rolled plate. The specimens were induction-heated to 1200°C at a rate of 2°C/s, and cooled to room temperature by Ar gas.

Torsion tests are ideally suited for the simulation of high temperature deformation processes characterized by large strains and high strain rates. Torsion tests were performed to an equivalent strain of 3.0 at temperatures ranging from 550°C (0.45 Tm) to 1050°C (0.75 Tm) at strain rates of 0.05 s⁻¹, 0.5 s⁻¹ and 2.0 s⁻¹. After the deformation, the torsion samples were water-quenched to room temperature within 1.0 sec to prevent any microstructural changes after deformation.

Torque and torsion angle measured during the torsion tests were converted to equivalent true stress, σ, and equivalent true strain, ε, using the following equations:

\[ \sigma = 3.3 \sqrt{T \pi R} \] .......................... (1)
\[ \varepsilon = \theta R / \sqrt{3L} \] .......................... (2)

\( T \) is the torque (N.m), \( \theta \) is the torsion angle (rad), \( R \) is the specimen radius (m) and \( L \) is the specimen length (m).

It is well known that the torsion specimens are deformed heterogeneously. The microstructure of the torsion specimen was therefore analyzed in the tangential plane at the external surface of the specimen. The microstructure of the tangen-

![Fig. 1. Optical micrograph of the microstructure after annealing at 1100°C for 300 seconds.](image-url)
tial plane was investigated by using an EBSD system mounted on a Field Emission (FE) Scanning Electron Microscope (SEM). Microstructural changes occurring during the cooling after the tests are known to be less pronounced for test carried out at low strain rates.\textsuperscript{21} The specimens selected for the analysis of the microstructural evolution were therefore those which had been deformed to an equivalent strain of 2.7 at a strain rate of 0.05 s\textsuperscript{-1}. Thin foils for TEM were prepared by mechanical grinding to a thickness of 130 μm. This was followed by twin-jet electro-polishing at 50 V in a solution of 5% HClO\textsubscript{4} + 95% CH\textsubscript{3}COOH at room temperature. The TEM specimens were observed in a FE-TEM operated at 200 kV.

3. Results

Figure 2 shows the Fe–Si binary equilibrium phase diagram calculated with the Thermo-calc software using the TCFE 6 data base. The measured dilatation curve, Fig. 3(a), and thermal expansion coefficient, Fig. 3(b), confirm that no phase transformation occurred during the heating of the specimen to 1200°C. It can be seen that the ferro- to paramagnetic transition occurred at 750°C, the Curie temperature of this alloy system.

Figure 4 shows the effects of deformation temperature and strain rate on the stress-strain curves. At all test temperatures, the stress-strain curves exhibited a peak stress. The specimen tested at 1050°C shows a low peak stress. After the peak, the stress decreases gradually with increasing strain. It should be noted that the equivalent stress of the specimens deformed at 850°C and above reached a steady state after the equivalent strain of 2.0. The magnitude of the peak stress increased with decreasing temperature and with increasing strain rate. In the case of deformations below 600°C, all specimens fractured during torsion test carried out with strain rates of 0.5 s\textsuperscript{-1} and 2.0 s\textsuperscript{-1}.

The occurrence of a single stress peak followed by a significant softening, as shown in Fig. 4, indicates that the steel undergoes dynamic recrystallization during hot deformation. It could be argued that the occurrence of a stress peak is due to the softening caused by deformation heat rather than dynamic recrystallization. In the present study, however, the possibility for the deformation heat as a main cause of dynamic softening was ruled out because the specimen temperature was carefully controlled to maintain constant value throughout the test. Furthermore, the presence of a peak stress was also observed for tests done at the highest temperature with the slowest strain rate where the contribution of deformation heat is negligible.

In order to get essential data from the stress-strain curves of Fe-2 mass% Si during hot deformation, the peak stresses were related to the strain rate and the deformation temperature by means of the hyperbolic sine function:\textsuperscript{22}

$$\dot{\varepsilon} = A \cdot (\sinh \alpha \sigma_p)^n \exp\left(-\frac{Q}{RT}\right)$$

where $A$, $\alpha$ and $n$ are temperature-independent material constants. $R$ is the universal gas constant (8.314 J/mol K), $T$ is the absolute temperature (K), $\sigma_p$ is the peak stress (MPa), $\dot{\varepsilon}$ is the strain rate (s\textsuperscript{-1}) and $Q$ is the activation energy for hot deformation (J/mol).

McQueen \textit{et al.}\textsuperscript{2} reported that the stress multiplier, $\alpha$, is an adjustable parameter that brings the factor $\alpha \sigma_p$ into the correct range to make the data points align in a parallel line for each temperature in $\ln \dot{\varepsilon}$ versus $\ln[\sinh(\alpha \sigma_p)]$ graphs. The stress exponent, $n$, which is given by the slope of the

\begin{align*}
\text{Fig. 2.} & \quad \text{Fe–Si binary phase diagram. The Si content and deformation temperatures are indicated by solid squares. The Si content of the steel used in the present study is indicated by dotted line. The test temperatures are indicated by the open squares in the diagram.} \\
\text{Fig. 3.} & \quad \text{(a) Dilatation ($\Delta L/L_0$) curves during heating and cooling and (b) Temperature- dependence of the thermal expansion coefficient ($d(\Delta L/L_0)/dT$) during heating.}
\end{align*}
data points in a ln{sinh($\alpha \sigma_p$)} plot, gives information about controlling mechanism of the hot deformation behavior in the deformation temperature range. The stress exponent is typically about 5 for pure metals, ceramics and many alloys over a wide temperature range and strain rate range.23) The activation energy for hot deformation has been reported to be close to the activation energy of lattice self diffusion, when the hot deformation behavior is controlled by dislocation climb. In the specific case of high temperature creep in fine grained materials at temperatures close to the melting temperature and at very low applied stresses, the deformation is controlled by the transport of vacancies through the grain interior from one grain boundary to other. This vacancy flux causes the grains to be elongated without dislocation slip or climb. The activation energy of the creep process also correspond to that of lattice self diffusion. The stress exponent is equal to 1 in this case and the deformation mechanism is generally referred to as Herring-Nabarro creep.23) 

In order to determine the parameters of the constitutive equation, the value of $\alpha$ parameter was first obtained by re-writing Eq. (3) in the following form:

\[
\ln \frac{\ln \ln \ln \left[ \sinh \left( \alpha \sigma_p \right) \right]}{\ln \left( \frac{Q}{RT} \right)} = n \ln \left[ \sinh \left( \alpha \sigma_p \right) \right] - A
\]

The plot of the logarithmic strain rate and ln{sinh($\alpha \sigma_p$)} is shown in Fig. 5(a). The value of $\alpha$ was obtained by fitting the data to Eq. (4) using the least squares fitting technique. An $\alpha$ value of 0.006 MPa$^{-1}$ and a $n$ value of 4.32 were found. This value for $n$ is consistent with a hot deformation controlled by the diffusional climb of dislocations. The value of $Q$, the activation energy of hot deformation, was cal-

\[ \text{Fig. 4.} \text{ Stress-strain curves for Fe-2 mass\% Si steel tested at various temperatures at strain rate of (a) 0.05 s}^{-1}, \text{ (b) 0.50 s}^{-1} \text{ and (c) 2.00 s}^{-1}. \]

\[ \text{Fig. 5.} \text{ Determination of the parameters in the Sellars-Tegart equation for the Fe-2 mass\% Si steel. (a) Stress versus strain rates, (b) Temperature versus flow stress and (c) ln{sinh($\alpha \sigma_p$)} versus ln Z.} \]
culated by rearranging Eq. (3) as follows:

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

(5)

As shown in Fig. 5(b), there is a clear linear relationship between \( \ln(\sinh(\alpha \sigma_p)) \) and the inverse absolute temperature. The slope of the straight data lines is equal to \( Q/nR \), which yields a value for \( Q \) equal to 276 kJ/mol. In order to calculate the constant \( A \), Eq. (5) was rearranged as follows:

\[ \ln \varepsilon + \frac{Q}{RT} = \ln A + n \ln[\sinh(\alpha \sigma_p)] = \ln Z \]  

(6)

\( Z \) is the Zener–Hollomon parameter, i.e. the temperature-corrected strain rate. Figure 5(c) shows a clear linear relationship between \( \ln Z \) and \( \ln[\sinh(\alpha \sigma_p)] \). This relationship is independent of the strain rate. The value of the constant \( A \) was determined to be 4.01 \times 10^{14}.

Figures 6(a) through 6(f) present the EBSD Image Quality (IQ) maps of the specimens deformed at various temperatures from 550°C to 1050°C, to an equivalent strain of 2.7 with a strain rate of 0.05 s\(^{-1}\). When the specimen was deformed at 550°C, a high density of LABs with misorientation between 3° and 15° (Green lines) was observed in the interior of the original grains, as shown in Fig. 6(a). The number of HABs with misorientation larger than 15° (Red lines) also increased in the microstructure during hot deformation. The HABs are aligned in specific directions, probably along the slip bands where extensive deformation occurs. Equiaxed crystallites surrounded entirely by HABs were also observed. One example of these equiaxed crystallites is indicated by a white arrow in the center of the micrograph in Fig. 6(a).

Figure 6(b) shows the IQ map of a specimen deformed at 650°C. Note that the magnification of the micrograph in Fig. 6(a) is higher than for the other micrographs in Fig. 6. The specimen deformed at 650°C is characterized by an inhomogeneous deformed structure. In the middle part of the micrograph, it can be seen that the area has a high density of both HABs and LABs. The density of these boundaries is much lower in the top and bottom part of the same micrograph. As will be discussed more in the following section, the deformed microstructure in these areas consists of a dislocation cell structure with a misorientation across the dislocation cell wall which is typically less than 2°. The microstructure in the central area of the micrograph consists of sub-grains surrounded by both HABs and LABs. Considering the original grain size of the specimen, it is expected that three to four original grains with an elongated shape are included in the micrograph after deformation. Gourdet and Montheillet reported that the evolution of the deformed microstructure in a single crystal was affected by the direction of the applied stress because the number of operating slip system varied with the deformation direction. Similarly, the inhomogeneous boundary distribution in Fig. 6(b) can be attributed to the evolution of different deformation microstructures depending on the orientation of the individual grains in the specimen.

Deformations at temperatures above 550°C resulted in the formation of equiaxed crystallites of which the size increased with increasing deformation temperature. Some crystallites were surrounded only by LABs. Other crystallites were surrounded by both LABs and HABs. It should be noted that HABs in the specimen deformed at 750°C were aligned in the direction of the maximum shear. It is very probable that some of the HABs were original grain boundaries. On the other hand, the HABs in the specimens deformed above 750°C were randomly distributed. This implies that new HABs were introduced during the deformation. In fact, this trend is not unique to the specimens deformed at temperatures above 750°C, but is already clearly visible in the deformed microstructure at 650°C. Furthermore, it must be noted that the ratio of LABs to HABs decreased with an increase of the deformation temperature. This observation cannot be explained by the standard theory for DRV as the only restoration mechanism for high SFE materials.

Figure 7 shows the variation in the distribution of boundary misorientation affected by the deformation at 850°C at a strain rate of 0.05 s\(^{-1}\). The fraction of boundaries with a certain misorientation was determined by counting the number of adjacent points having a specific misorientation within a 3 degree-misorientation interval. Adjacent points with misorientations less than 3 degree were not taken into account. Deformation to an equivalent strain of 0.2 resulted in an increase of densities in LABs with a misorientation typically less than 10°. The fraction of HABs was clearly reduced by further deformation to a strain of 0.5 due to the
drastic increase in a density of LABs. As the strain increased, the fraction of boundaries with a misorientation in the range of 8° to 15° clearly increased. Consequently, the misorientation distribution profile at an equivalent strain of 2.7, Fig. 7(d), can be described by the continuous decrease of the fraction of boundaries having a misorientation less than 15° and a homogeneous distribution of misorientation in the range of HAB, larger than 15°.

Figure 8(a) shows a TEM micrograph of the microstructure of the specimen deformed at 550°C at a strain rate of 0.05 s⁻¹. The deformation to an equivalent strain of 2.7 at 550°C resulted in a microstructure characterized by a formation of a well-developed dislocation cell structure and by the presence of a band of equiaxed crystallites similar to the one shown in Fig. 6(a). The cell structure shown in Fig. 8(a) is not equiaxed, but has a typical "block"-like structure with a high density of dislocation in the cell interior. The arrows in Fig. 8(a) indicate open cell walls. It is likely that new cell walls were formed in the cells. It can be seen that the size of the cell block and the directions of the cell wall are not uniform across the band of equiaxed crystallites. This implies that the local orientations of the two regions are different. Even though the exact misorientation could not be determined by diffraction pattern analysis, it is likely that the equiaxed crystallites were formed at boundaries between crystallites with a different crystallographic orientation. Figure 8(b) shows the dislocation structure in the interior of the cell structure. The dislocation pattern for two-beam contrast conditions and the direction of the dislocation Burgers vectors are indicated in the insets of Fig. 8(b). The dislocations in the cell structure are mainly screw type; they are slightly curved and jogged. Another striking feature in Fig. 8(b) is the presence of a large number of dislocation loops. The size of the loops is not uniform and, in the case of small loops, the loops are visible as small black spots in the micrograph, as shown in the inset indicated by A. In Fig. 8(b), some examples of larger dislocation loops are indicated by solid arrows. Dislocation cusps due to the presence of jogs on screw dislocations are shown in the inset and indicated by B. The presence of dislocation loops has frequently been observed in the microstructures of Fe–Si ferritic steels deformed at room temperature. The loops are known to arise from the pinch-off of dislocation dipoles formed by the presence of intermediate-sized jogs on moving screw dislocations. Figure 8(c) shows the microstructure of the equiaxed crystallites band observed in the central area of Fig.
8(a). The size of the crystallites ranged from 1 to 2 μm in diameter. It can be seen that the crystallites contain a high dislocation density comparable to that in the interior of the cell. From the diffraction pattern analysis, it is likely that the crystallites were more or less randomly distributed. The equiaxed crystallites were formed in an area of extensive deformation. The inhomogeneous microstructure and the presence of equiaxed crystallites are unique features of the microstructure after deformation at 550°C. This is also considered to be a rather an exceptional case for hot deformation behavior of Fe-2 mass% Si steel, since the deformation temperature of 550°C is lower than 0.5 Tm.

**Figure 9**(a) gives an overview of the microstructure obtained by TEM of the sample deformed at 650°C. The dislocations are organized in well-developed sub-grain structures. The dislocation density in the sub-grains is much lower than that in the interior of cell structure formed during deformation at 550°C. The dislocation distribution in the sub-grains varied from one sub-grain to the other. In general, the dislocations were more or less homogeneously distributed in the sub-grains. The dislocation distribution was not homogeneous when a new LAB had formed in the sub-grain. An example is indicated in the region indicated by black arrows. Figures 9(b) and 9(c) shows the dislocation structure in a sub-grain using different diffraction conditions. The dislocations in the sub-grain were mainly screw-type. There were jogs on the dislocation line. In contrast to the microstructure after deformation at 550°C, dislocation loops were rarely observed in the microstructure deformed at 650°C. It was reported that dislocation loops formed during deformation at room temperature disappeared during a subsequent heating to a temperature above 470°C. This observation was attributed to the migration of vacancies which allow for loop annihilation by dislocation climb. The absence of dislocation loops can, therefore, be explained by fast diffusion processes at 650°C, as the diffusivity of vacancies is high enough to eliminate the dislocation loops soon after their formation. A LAB consisting of an array of dislocations, denoted by A-B in Fig. 9(c), is present in the sub-grain. This LAB subdivides the sub-grain in two, slightly misoriented, parts. It is of interest to note that the density and the distribution of dislocations are different across the LAB. The region on the right has a higher dislocation density compared to that of the region on the left. The dislocation density in the left region is low in the vicinity of LAB, and it increases with increasing distance from the LAB.

This very characteristic dislocation density distribution is similar to the one originally introduced by Sandström et al. to illustrate the change in dislocation density close to moving HABs during dynamic recrystallization. The authors stated that the migrating grain boundary of a recrystallized grain will remove dislocations, and reduce the dislocation density to a very low value. After the movement of boundary, the dislocation density increases again by concurrent deformation. In the present case, it is likely that the LAB was moving in the sub-grain. A moving LAB can merge with a stationary boundary causing a change in the misorientation of this boundary, as suggested by Belyakov et al.

**Figure 10**(a) shows a TEM micrograph for the specimen deformed at 850°C. The microstructure is characterized by
the presence of sub-grains with a larger average sub-grain size and a lower sub-grain dislocation density compared to the specimen deformed at 650°C. This observation is in agreement with the reported relationship of sub-grain size and dislocation density in sub-grains.\textsuperscript{2,24} Open LABs, marked by arrows, were also observed in the microstructure. This observation implies that LABs formed in the sub-grain. Figure 10(b) shows the grain structure surrounded partly by HABs and partly by a LAB at right hand side top corner. This microstructure is characteristic for CDRx.

In the grain, a LAB consisting of a dislocation network (indicated by A), a boundary consisting of an array of dislocation (indicated by B) and dislocations forming a LAB (indicated by C) are also present. The misorientations caused by the presence of arrays of dislocation are, however, not so large since all dislocations are revealed by identical diffraction condition. The grain area, in particular in the boxed region, is characterized by a low density of straight dislocations free of jogs. The straight screw dislocation structure implies that diffusional processes at this temperature are fast enough to eliminate jogs by dislocation climb.

4. Discussion

4.1. Parameters for Constitutive Equation

The parameters of the constitutive equation for Fe-2% Si steel and other high Si alloyed steels reported in the literature are listed in Table 2.

The $\alpha$ value was determined to be 0.006 MPa$^{-1}$, in the present study. This deviates from the range of $\alpha$ value reported by previous studies.\textsuperscript{6,15–18,30} The $\alpha$ value is adjustable parameter which should be determined to make the data points in the $\ln \dot{\varepsilon}$ vs. $\ln \{\sinh(\alpha \sigma_b)\}$ plots aligned on a straight line. In the present study, the activation energy of hot deformation was assumed to be constant over the entire range of strain rate and test temperature.\textsuperscript{26} Therefore, the deviation of $\alpha$ value from the reported range can be attributed to the wider test temperature range used in the present study.

The value of $n$ in the present study was 4.32. This $n$ value is comparable with the values reported previously. Ani Shen \textit{et al.}\textsuperscript{30} reported a very different value for torsion tests in the range of 600°C–900°C. This could be attributed to the lower test temperature range than those of other reports. Belyakov \textit{et al.}\textsuperscript{31} reported that the mechanical behavior of metallic materials during high temperature deformation is usually determined by the diffusion-controlled climb of dislocations. Diffusion-controlled climb of dislocation is also the controlling mechanism during creep deformation at high temperatures. In the case of dislocation creep, the stress exponent has been shown to be in the range of 4–6.\textsuperscript{23} The activation energy, $Q$, for hot deformation is close to the activation energy for self diffusion. When a high stress is applied, plastic deformation is expected to occur mainly by the movement of dislocations on their slip plane. Dislocations interact with each other and form dislocation tangles during the deformation. Eventually, the dislocations are unable to move easily in their slip plane, and the screw dislocation segments move to other slip plane by cross slip and climb of edge dislocations and jogs on screw dislocations. In the case of dislocation climb, the vacancies in the crystal diffuse to the dislocation core and enable dislocations to climb to adjacent slip planes to resume glide deformation. As a result, the activation energy for hot deformation caused by dislocation climb is closely related to the activation energy for self diffusion. On the basis of the value of $n$, the hot deformation of ferritic Fe–Si steels can be said to be controlled by dislocation climb, with an activation energy for hot deformation similar to the activation energy for lattice self diffusion.

The activation energy of hot deformation was determined to be 276 kJ/mol. This value is slightly larger than the activation energy for self-diffusion in $\alpha$-Fe, 239.7 kJ/mol, reported by Buffington \textit{et al.}\textsuperscript{31} Stang et al.\textsuperscript{32} reported that the activation energy of high temperature creep in Fe-3 mass% Si steel was about 41.8–62.7 kJ/mol higher than that of self diffusion by experiments on the temperature dependence of elastic modulus. Treheux \textit{et al.}\textsuperscript{33} also reported that the activation energy of self-diffusion in Fe–Si alloys was in the range of 213–276 kJ/mol. Therefore, the activation energy determined in the present study can be regarded as being similar to that for self-diffusion in the Fe–Si alloy. This result is also in good agreement with the creep test results for $\alpha$-iron and its solid solution alloys of Karashima \textit{et al.}\textsuperscript{34}

This also implies that the hot deformation behavior of Fe–2 mass% Si alloy is controlled by diffusion-controlled climb of dislocations. In general, ferritic stainless steels alloyed with Cr have a higher activation energy, 397 kJ/mol, than Fe–Si ferritic steels.\textsuperscript{2} Ferritic stainless steels contain more than 10 mass% Cr. In addition, other elements such as Mo and Mn are added. These alloying elements increase the strength against deformation by solid solution hardening. Ferritic steels containing more than 2 mass% Si have a relatively low activation energy compared to ferritic stainless steels. The activation energy of ferritic Fe–Si steel is in the

Table 2. Parameter values for the constitutive equation for Fe–Si alloys. The alloy content is in mass%.

| Material          | $\alpha$ (MPa$^{-1}$) | $n$ | $Q$ (kJ/mol) | Test Temperature | Reference |
|-------------------|----------------------|-----|--------------|-----------------|-----------|
| $\alpha$-iron     | –                    | –   | 280          | 773 K–1 073 K   | Glover\textsuperscript{46} |
| Fe–2.05Si–0.003C–0.02Mn | 0.013    | 4.6 | 299          | 1 073 K–1 373 K | Calvillo\textsuperscript{47} |
| Fe–2.2Si          | 0.006               | 4.3 | 276          | 823 K–1 323 K   | Present work |
| Fe–2.7Si–0.06C    | 0.017               | 2.0 | 247          | 873 K–1 173 K   | Ani Shen\textsuperscript{30} |
| Fe–2.85Si–0.3C–0.45Mn | 0.017    | 4.3 | 334          | 923 K–1 173 K   | Uvira\textsuperscript{46} |
| Fe–3.0Si–0.03C–0.089Mn | 0.017   | 4.2 | 287          | 1 073 K–1 373 K | Akta\textsuperscript{30} |
| Fe–3.13Si–0.034C–0.02Mn | –        | 4.5 | 310          | 1 073 K–1 273 K | Inoue\textsuperscript{47} |
4.2. Dislocation Motion during Hot Deformation

TEM observations of the dislocation structure and the results of EBSD study show a clear dependence of the dislocation configuration on the deformation temperature. It is well known that DRv is mostly affected by the motion of dislocation, such as dislocation glide, climb and cross slip. In BCC metals, dislocation annihilation mechanism by climb of edge dislocations with opposite signs is not expected to have a significant effect on the dynamic recovery process, because the deformation is mainly controlled by screw dislocation. The climb mechanism, however, does contribute to the recovery process significantly by the following mechanisms. When the diffusion process is slow, the dislocations remain in a heavily jogged configuration. The jogs are formed by dislocation intersections during glide. The presence of jogs reduces the average mobility of screw dislocation by exerting a dragging force and the formation of dislocation dipoles on the moving dislocations. Cross slip is also expected to be less effective due to the non-pure screw character of the dislocations. Deformation at 550°C is also expected to have a significant effect on the dynamic recovery process, because the deformation is mainly controlled by screw dislocation. The climb mechanism, however, does contribute to the recovery process significantly by the following mechanisms. When the diffusion process is slow, the dislocations remain in a heavily jogged configuration. The jogs are formed by dislocation intersections during glide. The presence of jogs reduces the average mobility of screw dislocation by exerting a dragging force and the formation of dislocation dipoles on the moving dislocations.

4.3. The Mechanism for the Hot Deformation Behavior of Fe-2 mass% Si Steel

The occurrence of DRx in ferritic steels is still a matter of controversy as it has long been accepted that ferritic steels only undergo DRv during hot deformation. Due to their high stacking fault energies, dislocation climb and cross slip take place easily, leading to a mechanism for the reduction of stored energy that does not require recrystallization. Supporting this opinion, Inoue et al. reported the occurrence of DRv during hot deformation of a Fe-3 mass% Si steel during compression deformation. They reported that equiaxed sub-grains were formed by deformation at a very low strain rate, 2.2 × 10⁻⁵ s⁻¹, whereas a pan-caked grain structure was formed during deformation at a relatively high strain rate, 2.2 × 10⁻¹ s⁻¹. Based on this observation, they concluded that the microstructure resulted from DRv. Calvillo et al. investigated the hot deformation of a Fe-2 mass% Si steel in torsion. Although an unexpected softening phenomenon was observed, they insisted that the dynamic softening in 2 mass% Si steel was recovery-controlled because they found no evidence for recrystallization, i.e. the generation of new HABs, in their EBSD analysis of the microstructure.

In contrast, many studies have reported experimental evidence for the occurrence of DRx in ferritic steels. Baczynski et al. investigated the hot deformation behavior of high purity α-iron and two IF steels by means of torsion test. They found evidence for the occurrence of dynamic recrystallization by microstructural analysis. The recrystallized grains were uniformly distributed in the microstructure both at the original grain boundaries and in the interior of the original grains. This microstructure was very different from the necklace microstructure formed by DDRx where recrystallized grains should be observed only at the grain boundaries. Furthermore, they observed the formation of equiaxed grains at a strain of 2. The morphology of the equiaxed grains was not changed during the further deformation. Based on these observations, they concluded that the recrystallized microstructure of α-iron was results from the CDRx mechanism.

Oudin et al. investigated the hot deformation of Ti–IF steel in torsion. Based on their EBSD analysis, they reported that the number of new HABs segments increased with strain in large grained specimen during high temperature deformation. They also reported that the HABs segments could be found divorced from the initial grain boundaries. They explained this observation by the accelerated rotation of sub-grains due to differences in the crystallographic rotation field or to the activation of different slip systems from neighboring areas.

The results of present study show that the hot deformed microstructure of ferritic Fe-2 mass% Si steel consisted of equiaxed crystallites surrounded by large number of HABs, adding support to previous reports of dynamic recrystallization occurs in ferritic steel. In the present study, no evidence was found that the hot deformation microstructures of Fe-2 mass % Si steel at temperatures above 0.5 Tm developed by a GDRx mechanism. HABs were observed to be distributed homogeneously in the microstructure. These HABs are not determined activation energy also supports this explanation.
considered debris of the original microstructure because the amount of equivalent strain was not large enough to cause the pinch-off of the original grains.

Instead, it is proposed that the HABs observed in the present study are formed by a process of the gradual increase of the misorientation of the LABs. It was suggested that the insertion of lattice dislocations in LABs, which consist of dislocation arrays, leads to the reduced spacing of grain boundary dislocations.40 As lattice dislocations are gradually inserted into the boundaries, they become part of the grain boundary structure. The process of entrance of a lattice dislocation into a grain boundary and the formation of an extrinsic grain boundary dislocation has been described in the literature. At high temperature, where thermal activation is provided, these dislocations are annealed out either by dissociation into discrete structural grain boundary dislocation or by core gliding, enabling the continued absorption of lattice dislocation into the grain boundary. This process causes the rotation of the grain and HABs are formed by transition from LABs. The migration and merging of LABs can also contribute to the formation of HABs.29 In fact, many reports on the gradual increase of LAB misorientation strongly support the operation of CDRx mechanism in high SFE metals.29,40,44 In the present study, a gradual increase in the fraction of boundaries having misorientation in the range of 10° to 18° was observed in the specimen deformed at 850°C (Fig. 7). Therefore, the formation of HABs in Fe-2 mass% Si steel, at temperatures above 0.5 Tm, can be attributed to the gradual increase of misorientation of LABs by the absorption of dislocations. The formation of equiaxed crystallites in the specimen deformed at 550°C, on the other hand, can be attributed to the GDRx mechanism, but this should be considered to be exceptional as it would result from a deformation at low temperature.

5. Conclusions

The hot deformation behavior of ferritic Fe-2 mass% Si steel was investigated by means of torsion tests in the 0.45 Tm to 0.75 Tm temperature range. The microstructure was analyzed by EBSD and TEM.

Dynamic softening with a single stress peak was found to occur in the entire test temperature range. The parameter values in the hyperbolic sine type constitutive equation of Fe-2 mass% Si steel were as follows: α = 0.006 MPa⁻¹, n = 4.32, Q = 276 kJ/mol. The activation energy for hot deformation is close to that for α-iron lattice self diffusion, indicating that hot deformation of Fe-2 mass% Si steel occurs by a thermally-activated, diffusion-controlled dislocation climb mechanism.

The hot deformed microstructure of ferritic Fe-2 mass% Si steel consisted of equiaxed crystallites surrounded by a high number of HABs. The HABs were homogeneously distributed in the microstructure. In addition, a gradual increase of the number of boundaries having a misorientation in the range of 10° to 18° was observed in the specimen deformed at 850°C. On the basis of these observations, the formation of HABs in Fe-2 mass% Si steel deformed at temperatures above 0.5 Tm was attributed to a gradual increase of the misorientation of sub-grains brought about by the absorption of dislocations.

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