Brittle-to-ductile Transition in Nickel-free Austenitic Stainless Steels with High Nitrogen

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The brittle-to-ductile transition (BDT) behaviour in nickel-free austenitic stainless steel with high nitrogen was investigated. Fall-weight impact tests revealed that Fe–25mass%Cr–1.1mass%N austenitic steel exhibits a sharp BDT behaviour in spite of an fcc alloy. The aspects of plastic deformation after the impact tests indicate that the BDT observed in this austenitic steel is induced by poor ductility at low temperatures as is the same as that in ferritic steels. In order to measure the activation energy for the BDT, the strain rate dependence of the BDT temperature was examined by using four-point bending tests. The weak dependence of the BDT temperature on the strain rate was observed. The Arrhenius plot of the BDT temperature against the strain rate elucidated that the activation energy for the BDT of Fe–25mass%Cr–1.1mass%N is much higher than that of low carbon ferritic steels. The origins of such distinct BDT behaviour and its large value of the activation energy in this high-nitrogen steel are discussed in terms of the reduction of dislocation mobility at low temperatures due to the interaction between glide dislocations and nitrogen solute atoms.

KEY WORDS: dislocation; fracture; activation energy; dislocation mobility.

1. Introduction

Austenitic stainless steels are widely used in line pipes, chemical storages, electric power plants and so on, due to their eminent mechanical properties and resistance to wear. Solution nitriding of austenitic stainless steels has attracted attention in order to reduce the amount of nickel used. Solute nitrogen is not only a strong austenite-former but also enhances wear resistance and mechanical properties such as strength and toughness, so that it is expected to be used as a substitute of nickel to form austenitic stainless steels with high strength and toughness. However, one of the problems for the usage of high nitrogen steels is that they exhibit a brittle-to-ductile transition (BDT) which is not basically seen in fcc alloys.1–3) There are several factors suggested to explain the onset of the BDT behaviour in high-nitrogen stainless steels as follows.

(1) deformation twinning;4) a crack advance is induced by collisions of deformation twins.
(2) ε-martensite transformation;5) a cleavage crack is induced by the collisions of ε-martensite.
(3) slipping-off induced by solute nitrogen;1,6) dislocations keep gliding on a single {111} plane, which leads to the initiation of micro-cracks to fracture.
(4) pile-up of planer dislocations against grain boundaries;7) the rare opportunity of cross-slip induces pile-up dislocations against grain boundaries, leading to high stress intensity around the grain boundaries which causes fracture. It is still ambiguous, however, why high-nitrogen austenitic steels exhibit the BDT behaviour.

The BDT behaviour is generally seen in bcc metals, covalent crystals and ceramics in which dislocations need to overcome high Peierls barriers to move. Pioneering works to understand the fundamental mechanism behind the BDT were performed by using several sorts of single crystals.8–12) They revealed that the BDT temperature is strain rate dependent and also that the one of the most essential controlling process of the BDT is dislocation gliding, which was accepted from the fact that the values of activation energy obtained from the strain rate dependence of the BDT is close to those for dislocation gliding. Table 1 shows the agreement of activation energies for the BDT with those of dislocation gliding in various crystalline materials.9,13–15) It suggests the analogy that the BDT behaviour in high-nitrogen austenitic steels is also controlled by the dislocation gliding process.

Table 1. Comparison between the activation energy for the BDT and dislocation gliding.

| Materials | Activation energy (eV) |
|-----------|-----------------------|
|           | BDT                  | Dislocation gliding |
| Mo9)      | 0.49                 | 0.49                |
| Ge13)     | 1.54 ± 0.05          | 1.58                |
| Si (n-type)16) | 1.6 ± 0.1       | 1.7                 |
| Si (intrinsic)14) | 2.1 ± 0.1   | 2.2                 |
| Sapphire15) | 3.2 ± 0.2       | 3.6                 |
In the present study, the mechanism behind the BDT in nickel-free high-nitrogen austenitic steels will be discussed, focussing on a dislocation activity. The temperature dependence of 0.2% proof stress, which has a close relation to dislocation activities, was firstly measured. Then, the BDT behaviour was investigated by performing impact tests with miniature size specimens. The activation energy was obtained from the strain rate dependence of BDT temperatures by performing four-point bending tests with three different crosshead speeds. The activation volume was also measured by strain-jump tests.

2. Experimental

The chemical composition of the specimen employed is shown in Table 2. 1.1 mass% nitrogen was added into Fe–25 mass%Cr by using solution nitriding process at 1473 K for 72 ks in nitrogen gas under the pressure of 0.1 MPa. The specimen was water-cooled not to cause austenite decomposition and induce nitride precipitates during the cooling. The solution nitrided steel was subjected to an isothermal heat treatment in a dual phase region at 1173 K for 0.3 ks. Finally, the isothermally heat-treated steel was reaustenised at 1473 K and then quenched. Tensile tests were performed at various temperatures under the initial strain rate of $1.8 \times 10^{-4}$ s$^{-1}$. Impact absorbed energy was measured by using an instrumental impact tester (TANAKA, TM-H-150) to highlight the BDT behaviour in high nitrogen steels. Apparent fracture toughness, $K_{Q}$, was obtained by using four-point bending tests at various temperatures with the following equations:

\[
K_{Q} = \sigma_{\text{frac}} \sqrt{\pi a F_{I}(\xi)}, \quad \xi = \frac{a}{h},
\]

\[
F_{I}(\xi) = 1.122 - 1.40\xi + 7.33\xi^{2} - 13.08\xi^{3} + 14.0\xi^{4}, \quad \ldots \quad (1)
\]

where $\sigma_{\text{frac}}$ denotes the fracture stress, a and h denote the notch length and specimen height, respectively. The inner and outer beam spacing are 4 mm and 10 mm, respectively. Activation energy was obtained from the strain rate dependence of the BDT temperatures.

In order to estimate the stability of austenite phase, the temperature at which 50% martensite is produced under the action of a true strain of 0.30, $\text{Md}_{\text{30}}$, and the martensite-start temperature, $\text{Ms}$, were calculated by the following equations for austenitic steels:\(^{16}\)

\[
\text{Md}_{\text{30}}(\degree C) = 497 - 462 (%C+N) - 9.2 (%Si) - 8.1 (%Mn) - 13.7 (%Cr) - 20 (%Ni) - 18.5 (%Mo) \ldots \quad (2)
\]

\[
\text{Ms}(\degree C) = 502 - 810 (%C) - 1.230 (%N) - 13 (%Mn) - 30 (%Ni) - 12 (%Cr) - 54 (%Cu) - 46 (%Mo) \ldots \quad (3)
\]

The apparent values of $\text{Md}_{\text{30}}$ and $\text{Ms}$ were calculated to be $-353\degree C$ and $-1151\degree C$ for the material used in this study, respectively, which means that the austenitic phase of the specimens used in the present steel is very stable during the series of mechanical tests at any temperatures employed.

3. Results

Figure 1 shows true stress-strain curves from Fe–25Cr–1.1N steel obtained by tensile tests at 150 K, 250 K and 300 K. Work hardening rates are also exhibited at the right top of the figure. The specimens deformed at 150 K and 250 K fractured at the strain of approximately 0.02. The tensile test at 300 K was also terminated at the strain of 0.02 in this case when the strain gage attached was flaked from the specimen surface. Nominal tensile strain of the specimen fractured at room temperature reaches around 0.5, which indicates the stress-strain relation of this high nitrogen steel exhibits remarkable temperature dependence as reported previously.\(^{7,17,18}\) Figure 2 shows the temperature dependence of yield stress (0.2% proof stress) obtained from Fig. 1, where that from SUS316L\(^{19}\) is also plotted for comparison. Yield stresses in both steels are increasing with the decrease in temperature. Note, the temperature dependence

**Table 2.** Chemical composition of the specimen used.

| C | Si | Mn | P | S | Cr | N | Fe |
|---|---|---|---|---|---|---|---|
| 0.002 | < 0.01 | < 0.01 | < 0.005 | 0.0004 | 25.11 | 1.1 | bal. |

mass%
in Fe–25Cr–1.1N steel is much stronger than that in SUS316L steel. If the values of Fe–25Cr–1.1N are linearly extrapolated to 77 K, the increases in 0.2% proof stress between 300 K and 77 K in SUS316L and Fe–25Cr–1.1N are estimated to be 130 MPa and 700 MPa, respectively. Such remarkable temperature dependence of yield stress observed in this high nitrogen steel should induce embrittlement at low temperatures, i.e., the onset of the BDT in spite of austenitic steels.

Next, to investigate the BDT behaviour of the Fe–25Cr–1.1N steel, impact tests were performed at various temperatures between 120 K and 350 K. Figure 3 shows SEM images obtained from the specimens fractured by the impact tests at 176 K, 273 K and 336 K. In Fig. 3(a), at 176 K, the crack extended nearly straight from a notch root with little sign of plastic deformation, indicating brittle behaviour. In the specimen tested at 273 K, as shown in Fig. 3(b), zigzag crack extension occurred, partially indicating plastic deformation along the crack wake. Figure 3(c), at 336 K, demonstrates remarkable plastic deformation and crack blunting instead of crack extension. Enlarged images of fracture surfaces in Figs. 3(g)–3(i) indicate the fracture mode changes from inter-granular to trans-granular as the test temperature increases. The specimen used in this study does not contain enough amount of carbon to prohibit inter-granular fracture. The SEM images exhibit a strong temperature dependence of crack-tip plasticity in Fe–25Cr–1.1N steel, from which the occurrence of sharp transition is expected in the values of absorbed impact energy.

Figure 4 shows the absorbed impact energy as a function of temperature obtained from the Fe–25Cr–1.1N steel, where that from SUS316L measured in this study are also indicated for comparison. SUS316L does not exhibit a BDT behaviour in the present temperature range, and instead of that, there appears slight decrease in absorbed energy with increasing temperature, approximately 200 kJ/m² between 150–300K. It is regarded to be the reverse temperature dependence of absorbed energy often seen in the upper-shelf of BDT curves. It suggests that the BDT temperature of SUS316L is much lower than 77 K which is the minimum test temperature obtainable in the present study. On the other hand, the absorbed energy obtained from Fe–25Cr–1.1N abruptly changes around at 270 K, exhibiting a sharp BDT as is commonly observed in ferritic steels.

The BDT behaviours are generally seen in materials such as bcc metals, semiconductors and ceramics, which have been considered to be due to the strong temperature dependence of the dislocation glide. Pioneering works using those single crystals8–12 clarified that the BDT is a process controlled by the dislocation glide since the values of activation energy obtained from the strain rate dependence of the BDT temperature is nearly the same as those for the dislocation glide, as was seen in Table 1. It suggests that, also in Fe–
25Cr–1.1N steel, the activation energy for the BDT should be regarded as that for the dislocation glide. In order to know the activation energy for the BDT in this steel, strain rate dependence of the BDT temperature was examined by performing four-point bending tests.

Figures 5(a), 5(b) and 5(c) show temperature dependence of apparent fracture toughness, \( K_0 \), with the crosshead speeds of 0.05 mm/min, 0.5 mm/min and 5 mm/min, respectively. \( K_0 \) exhibits a drastic change with increasing test temperatures; when temperature is raised from 77 K, \( K_0 \) values increase at first and take the highest values around the temperatures of 160–170 K, and then decrease with increasing temperature. Even though the values of apparent fracture toughness are nearly the same, large plastic deformation was seen beneath the notch at higher temperatures. Open symbols indicate the case that the distance of plastic deformation expanded from the notch root was longer that the notch depth, observed from side surfaces. Although some experimental scattering is observed in all the figures, the BDT temperature is defined as the temperature with the highest value of \( K_0 \) in this study, which is indicated by a dashed line in each figure of Figs. 5(a), 5(b) and 5(c).

The BDT temperatures at the crosshead speeds of 0.05 mm/min, 0.5 mm/min and 5 mm/min were determined to be 160 K, 160 K and 165 K, respectively, indicating a weak dependence of BDTT on the strain rate. The BDT temperature experimentally obtained at the crosshead speeds of 0.05 mm/min and 0.5 mm/min are the same whereas the BDT temperature should increase with increasing the crosshead speed. It is due to the experimental scatter of the \( K_0 \) in this study. The effect on the activation energy of the scatter will be briefly discussed later. Note here that the decrease in \( K_0 \) observed in the right side of the dashed lines does not mean the actual decrease in toughness but is due to the large amount of plastic deformation around the notch root, where the condition for small scale yielding is not satisfied at all.

The relationship between the BDT temperature and strain rate is empirically given by the following equation:

\[
\dot{\varepsilon} = \varepsilon_0 \exp\left(-\frac{Q_E}{kT_{BDT}}\right) \quad \text{(4)}
\]

where \( \varepsilon_0 \) is a prefactor, \( k \) and \( T_{BDT} \) are Boltzmann constant and the BDT temperature, respectively. \( Q_E \) is the activation energy for the controlling process of the BDT, i.e. the dislocation glide. Arrhenius plot of the strain rate against the BDT temperature obtained from Fig. 5 is shown in Fig. 6 with the result from low carbon steel. Equation (4) is valid in the case of specimens with the uniform grain size. The grain size of the material used is approximately 50–80 \( \mu m \). It is reported high nitrogen steels have a large Hall-Petch

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Fig. 5. Apparent fracture toughness as a function of temperature from Fe–25Cr.11N. Crosshead speeds were set to be 0.05, 0.5, 5 mm/min in (a), (b) and (c), respectively. The temperature at which the apparent fracture toughness indicates the highest was defined as the BDT temperature. Open symbols indicate the area of plastic deformation was expanded more than the length of notch depth.

Fig. 6. Arrhenius plots obtained from the deformation dependence of the BDT temperature. The activation energy for Fe–25Cr–1.1N was found to be \( 1.6 \pm 0.9 \) eV, which is much higher than that for low carbon steel obtained in previous study.
coefficient so that the variation of the grain size strongly influences the precision of the measurement of activation energy. In the present study, Fig. 2 shows that plots of 0.2% proof stress versus temperature are on a single line, indicating the variation of strength due to that of grain size is negligible so that it is reasonable to use Eq. (4). The slope from Fe–25Cr–1.1N gives the activation energy of 1.6 ± 0.9 eV, where its standard error is used as the margin of error. Although the margin of error is relatively large and the value itself is provisional, it is to be stressed here that the deformation rate dependence of BDT temperature is much weaker than that in ferritic steels, which induces much larger value of the activation energy for the BDT in Fe–25Cr–1.1N than ferritic steels of 0.3 eV.

Since the Peierls barrier for dislocation motion in fcc crystals is negligibly small compared to that of screw dislocations in bcc crystals or semiconductors with strong covalent bonding, the high activation energy of the BDT in the present case is not due to the existence of such elements as carbon which induces the difficulty for dislocations to glide at low temperatures and the BDT behaviour, will be discussed next.

4. Discussion

Among many factors influencing the plastic deformability in crystalline materials, the effect of inter-granular fracture on the BDT temperature will be discussed first. The criterion for brittle fracture with plain strain condition is derived as follows in Griffith theory:

\[ K_{\text{IC}} = \frac{2E\gamma}{1-\upsilon^2}k_D, \]

where \( E \), \( \gamma \) and \( k_D \) are Young’s modulus, surface energy, Poisson’s ratio and local stress intensity factor due to dislocations. The first term in the right-hand in Eq. (5) is relating to atomic biding energy while the second term is relating to the ability to accommodate stress concentration at a crack tip. The sign of \( k_D \) is negative when dislocations are emitted from the crack tip so that \( K_{\text{IC}} \) increases as the number of dislocations increases. Here, \( \gamma \) is given as follows in cases of trans-granular and inter-granular fractures:

\[ \gamma = \gamma_e : \text{trans-granular fracture} \]
\[ \gamma = \gamma_s : \text{inter-granular fracture}, \]

where \( \gamma \) is trans-granular fracture surface energy and \( \gamma_e \) is grain boundary energy. Equation (6) suggests if grain boundary energy increases but dislocation mobility is not changed, fracture toughness is decreased and the fracture mode would change from trans-granular to inter-granular. The material used in this study exhibits inter-granular fracture due to the lack of elements such as carbon which increases the value of grain boundary binding. It is expected that the addition of a little amount of carbon would increase the BDT temperature but still the material must exhibit the BDT behaviour. The inter-granular fracture is not due to the existence of nitrogen atoms. Therefore, the inter-granular fracture seen in this study is not essential for the onset of the BDT in this material used. Nitrogen is considered to affect the second term in right-hand of Eq. (5), i.e., stress accommodation term due to plastic deformation since the strong temperature dependence of 0.2% proof stress suggests the dislocation velocity is strongly influenced by dislocation mobility as shown in Fig. 2. Therefore, the most essential factor to be considered next is the primitive process of dislocation gliding, particularly in materials with high values of Peierls barrier for dislocation motion.

The value of activation energy for the BDT has been investigated in various materials which have high Peierls barriers. Figure 7 shows the relation between the activation energy for the BDT and the BDT temperature for those materials. Here, the activation energy obtained from the present study of Fe–25Cr–1.1N is overlaid. A regression line in Fig. 7 is drawn from the data excluding the plot from Fe–25Cr–1.1N, fixing one edge of the line with the origin of the graph. It exhibits a good liner relation between the activation energy and the BDT temperature, where the plots from Fe–25Cr–1.1N are deviating from the regression line.

There are two points to be emphasised here to understand the deviation from the regression line;

1. Fe–25Cr–1.1N austenitic steel exhibits a very large value of the activation energy compared to that of ferritic steels.

2. In spite of the large value of activation energy, Fe–25Cr–1.1N exhibits a very low BDT temperature such as 160 K. Materials with nearly the same activation energy such as n-type silicon exhibit the BDT temperature around 850 K.

Firstly, in order to discuss the large value of the activation energy for the BDT in Fe–25Cr–1.1N, we consider the thermally activated process of the dislocation glide. The dislocation velocity, \( v \), is generally given by:

\[ v = v_0\tau_e^m\exp\left(-\frac{Q_e}{kT}\right), \]

where \( v_0 \), \( \tau_e \), and \( T \) are pre-factors, the effective stress, and the absolute temperature, respectively. Figure 7 shows the relation between the activation energy and the BDT temperature. The results of present study are deviated from the regression line drawn from the other points. Other data are from: Si, GaAs, Ge, α-Al₂O₃, Mo, SiC, TiAl, diamond, Fe–3%Si, V, Fe, and Fe–25Cr–1.1N. The deviation is due to the presence of Peierls barriers for dislocation motion.
temperature, respectively. \(Q_d\) is the activation energy for dislocation gliding, which is consider to be the same as \(Q_S\) in Eq. (4). Here, \(Q_d\) is a function of \(\tau_c\), applied shear stress, \(\tau_c\), athermal stress, \(\tau\), and activation volume, \(v^*\):

\[
Q_d = F - \tau_v v^* \quad \text{(8)}
\]

\[
\tau_v = \tau_a - \tau \quad \text{(9)}
\]

where \(F\) is Helmholtz free energy for the dislocation glide, the amount of which depends on the nature of the short range barrier. Here, \(v^*\) is given by:

\[
v^* = M_f \frac{kT\ln v_0}{\partial \sigma} \quad \text{(10)}
\]

where \(\sigma\) is the applied tensile stress and \(M_f\) is Taylor factor of 3 in fcc metals. The activation volume was obtained in this study by performing strain jump tests from \(\dot{\varepsilon} = 1.0 \times 10^{-4}\) to \(1.5 \times 10^{-3}\). The activation volume at 150 K was found to be approximately \(25b^3m^3\) (b: the absolute value of the Burgers vector) which is the same level of the value with that for dislocations to overcome Peierls barriers in ferritic steels. This fact and the value of \(Q_d\) in Fe–25Cr–1.1N is larger than that in ferritic steels suggest that either \(F\) is much larger or \(\tau_c\) is much smaller than those in ferritic steels or both happens. Tomota et al. indicated nitrogen increases internal stress, which suggests the decrease in \(\tau_c\), leading to the increase in \(Q_d\) in Eq. (8) while the possibility of the large value of \(F\) in Eq. (8) could be explained by a strong interaction between dislocations and solute nitrogen.

Several models have been suggested for solid solution hardening in fcc metals. Gavrlijuk et al. explained the temperature dependence of yield stress in Cr18Ni16Mn10N0.4 steels basing on Seeger’s theory. He showed that the tendency of the temperature dependence itself does not depend on the existence of nitrogen in Cr18Ni16Mn10 steels, and suggested that the strong temperature dependence of yield stress below 100 K is due to the increase in the stress for screw dislocations to form vacancies when they glide as expected in Seeger’s theory. They assumed that the increase in nitrogen content would reduce the stacking fault energy to form wider stacking faults, which needs higher applied stress for split dislocations to constrict to form the vacancies. Bryrnes et al. and Grujicic et al. also showed the strong temperature dependence of nitrogen doped steels. They suggested the thermally activated process in dislocation gliding is the interaction between dislocations and short-range ordering atoms.

Marukawa has recently suggested a modified model of solid solution hardening for fcc metals, taking plural atoms into account in the interaction process. The assumption is that plural atoms simultaneously dedicate to a thermally activated process of dislocation gliding. One could suggest from that model that such contribution of plural solute nitrogen increases the activation energy of the BDT, which is because the value of activation energy in the model is proportional to the number of atoms which are simultaneously involved in the thermally activated process. It is, therefore, speculated that plural nitrogen atoms play a role as strong short-range barriers against dislocation gliding in Fe–25Cr–1.1N, which gives the large value of \(F\) in Eq. (8). There are few experimental data of the activation energy for dislocation gliding in fcc alloys due to the difficulty of the measurement. Suzuki and Kuramoto measured the activation energy of dislocation gliding in Cu–2%Al alloys and found it to be 0.7 eV at room temperature, which corresponds to that without effective stress. Kozlov et al. showed in austenitic steel of Cr18Ni25M06N0.7 that enthalpy of the binding between nitrogen atoms and dislocations drastically increases from 0.6 eV to 1.18 eV with nitrogen content from 0.065 to 0.65 mass%. Interstitial nitrogen in austenitic steels interacts with dislocations stronger than substitution atoms such as aluminum in copper. The strong interaction of interstitial nitrogen should induce to the increase \(F\) in Eq. (6).

The second problem to be discussed is why the BDT temperature of Fe–25Cr–1.1N is so low in spite of its high activation energy nearly the same value as that in n-type silicon. The transition of the fracture mode from trans-granular to inter-granular increases the BDT temperature, which induces the shift of the regression line in Fig. 7 to right, so that the deviation of the plot from Fe–25Cr–1.1N from the regression line cannot be explained by the inter-granular fracture. Therefore, it is necessary to focus on the dislocation activity in this material. The dislocation velocity is given by the same form of Eq. (7) irrespective of crystal structures, i.e., fcc or diamond structure as far as the dislocation gliding is the thermally activated process. Effective stress on dislocations in n-type silicon, which gives the activation energy of 1.6 eV, is around 20 MPa while m in Eq. (7) is close to 1. Those values are considered to be nearly the same as those in fcc metals. The fact that the BDT temperature in Fe–25Cr–1.1N is much lower than that in n-type silicon suggests that the pre-exponential factor, \(v_0\), of Fe–25Cr–1.1N in Eq. (7) is much larger than that of n-type silicon.

Suppose that short-range barriers which dislocations overcome exist with their spacing of \(d\) as shown in Fig. 8. The duration necessary for a dislocation to overcome one short range barrier, \(\Delta t\), is proportional to:

\[
v_0^d\exp\left(\frac{Q_d}{kT}\right) \quad \text{(11)}
\]

where \(v_0\) is a factor with respect to Debye’s frequency. Since the number of barriers along the distance of \(L\) is given by \(L/d\), the mean velocity for a dislocation to travel the distance of \(L\) is proportional to:

\[
dv_0^d\exp\left(\frac{Q_d}{kT}\right) \quad \text{(12)}
\]

Debye’s frequency of silicon is slightly larger than that of...
iron but they are the same order, which indicates that \( d \) in Fe–25Cr–1.1N should be much larger than that in silicon. Here, \( d \) in n-type silicon should be the same order with the Burgers vector, because short range barrier to overcome in silicon is Peierls barriers. On the other hand, spacing \( d \) in Fe–25Cr–1.1N must be much larger than Burgers vector, where dislocations interact with nitrogen atoms as seen in Fig. 8. The larger value of \( d \) in Eq. (12) in Fe–25Cr–1.1N leads to the smaller value of \( T \), supposing that the average dislocation velocities in Fe–25Cr–1.1N and n-type silicon are the same at their BDT temperatures: 150 K for Fe–25Cr–1.1N and 850 K for n-type silicon. Thus, in spite of nearly the same values of their activation energies, dislocations in Fe–25Cr–1.1N can move fast enough at low temperatures at which dislocations in n-type silicon cannot, because of the difference in average spacing of short range barriers.

5. Conclusions

The brittle-to-ductile transition behaviour in nickel-free austenitic steel with high nitrogen was studied performing tensile tests and miniature size four-point bending tests. The summary of the obtained results is as follows:

1. Temperature dependence of 0.2% proof stress in Fe–25Cr–1.1N is much stronger than that in SUS316L.

2. No apparent signs of plastic deformation were seen along the crack wake in specimens tested at temperatures among the lower-shelf in the curve of the absorbed energy.

3. The activation energy for the BDT temperature in Fe–25Cr–1.1N was much larger than that in ferritic steels. It suggests that the difficulty of dislocation gliding in Fe–25Cr–1.1N at low temperatures induces the BDT.

4. It is speculated that solute nitrogen atoms increase in Helmholtz free energy or/and internal stress for dislocation gliding, which leads to the large value of the activation energy for the BDT.

5. The difference of the thermally activated process for dislocation gliding between fcc and diamond structure can explain the reason why the BDT temperature in Fe–25Cr–1.1N is much lower than that in n-type silicon even though they have nearly the same activation energy.

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