Dynamic $\gamma \rightarrow \alpha$ Transformation during Hot Deformation in Iron–Nickel–Carbon Alloys

Hiroshi YADA, Chun-Ming Li1 and Hiroshi YAMAGATA2

Shizuoka Institute of Science and Technology, Fukuroi-City 437-8555 Japan. 1) Now with Max-Planck Inst. for Metal Res., 70174 Stuttgart, Germany. 2) Yamaha Motor Co. LTD., Iwata-City, 438-8501 Japan.

(Received on September 2, 1999; accepted in final form on October 13, 1999)

Fe–6mass%Ni–(0.0008–0.29)mass%C alloys were hot-deformed in torsion at 600–720°C (above the cooling transformation start temperatures $A_{1p}$) after austenitization. An in-situ X-ray diffraction study revealed that $\gamma \rightarrow \alpha$ transformation occurred during deformation in a wide range of condition, even above $A_{1p}$ (paraequilibrium $\gamma \rightarrow \alpha$ transformation temperature). Corresponding to this transformation, apparent decrease in deformation stress from that expected for austenite was observed. Microstructural study of the specimens quenched after the deformation showed that a large amount of fine-grained ferrite was formed due to the deformation. The analysis of deformation stress and chemical driving-force of the transformation indicated that the transformation occurred in order to reduce the total energy of deformed material since the deformation of energy of $\alpha$ was revealed to be considerably smaller than that of $\gamma$ and the amount of deformation energy saved by the transformation was shown to be much greater than the chemical energy consumed by the transformation at the tested temperatures.

KEY WORDS: low carbon steels; nickel steels; hot deformation; grain refinement; thermomechanical heat treatment; strain-induced transformation.

1. Introduction

One of the authors has reported that a large amount of ultra fine-grained ferrite of 1 to 4 $\mu$m diameter was formed in various low carbon steels, when the austenite was heavily hot-deformed at the temperature range around $A_{1p}$ or above and then rapidly quenched.1–4) This finding has suggested a new possibility of developing high performance steels without the addition of special alloying elements.

Evidence has suggested that the ferrite be formed during hot deformation. First, the deformation stress observed during the hot deformation was much lower than that expected for austenite, and the decrease in deformation stress was observed to be proportional to the fraction of ferrite formed. The deformation stress of low carbon ferrite is generally accepted to be considerably smaller than that of austenite compared at the same temperature. Similar reduction of deformation stress in the $\gamma \rightarrow \alpha$ two phase region in low carbon steels was also reported by other authors.5) Second, even very high quenching rates could not suppress the formation of the ferrite, which amounted to over 90% in volume fraction when large strain had been applied to a specimen. Third, the ferrite was unstable above the equilibrium start temperatures of proeutectoid ferrite, $A_{1p}$, and decreased in volume fraction with time during isothermal holding following deformation, while it was stable below $A_{1p}$ and its grain size increased during isothermal holding.

Since recovery and recrystallization during deformation are called dynamic recovery and recrystallization, respectively, we will call this transformation “direct transformation (DT)”. Formation of ferrite above $A_{1p}$, however, posed a serious question about the driving force of the transformation. Furthermore, in this transformation thermodynamical transformation start temperature should have been the temperature at which the free energy of the bcc phase is equal to that of the fcc phase, $T_{fi}$, which was considerably lower than $A_{1p}$. This is because long range diffusion of carbon could not substantially have accompanied the DT since the reaction was shown to be insensitive to strain rates up to $250\,s^{-1}$ and therefore the partition of carbon between austenite and ferrite was not expected substantially. If we admit this inference, the driving force for $\gamma \rightarrow \alpha$ transformation is considered to be negative in most of the above experiments.

For explaining this problem, it was proposed that the chemical driving force would have been supplied by mechanical work during plastic deformation. The energy of plastic deformation was calculated to be more than 20 times as large as the chemical driving force required to cause DT in some of the above experiments.5,6) Furthermore, it was shown that the saved deformation energy due to the decrease in deformation stress during DT was estimated to be 9 to 14 times as large as the chemical energy supposed to be dissipated by the transformation at the tested temperatures. This analysis indicated that DT occurred in order to reduce the total energy of deformed material.

No direct evidence, however, has hitherto been obtained to prove DT except the reductions of deformation stress, which the dynamic recrystallization of austenite could also bring about. In this connection an in-situ X-ray diffraction
study during hot deformation has been undertaken in low carbon Fe-Ni alloys. The in-situ X-ray diffraction technique had been successfully applied to find dynamic recrystallization of aluminum.7) The experimental results of the technique had been successfully applied to find dynamic recrystallization of aluminum. The experimental results of the technique had been successfully applied to find dynamic recrystallization of aluminum. The experimental results of the technique had been successfully applied to find dynamic recrystallization of aluminum. The experimental results of the technique had been successfully applied to find dynamic recrystallization of aluminum. The experimental results of the technique had been successfully applied to find dynamic recrystallization of aluminum.

2. Experimental

2.1. Materials

The compositions of the alloys used in this study are shown in Table 1. Nickel was added to reduce transformation temperature so as to make high temperature X-ray studies easier. It also enhances the hardenability of the alloys. Table 2 gives the transformation temperatures of the alloys. Equilibrium transformation temperatures were calculated by using Thermo-Calc. In this alloy system transformation during cooling was known to start from their paraequilibrium transformation start temperature, $A_{1}^{p}$, not from orthoequilibrium transformation start temperature, $A_{1}^{o}$. Experimental cooling transformation start temperatures, $A_{15}$, and transformation heating finish temperatures, $A_{13}$, and a martensite start temperature, $M_s$, were adopted from preceding work,12) in which alloys of the compositions close to the present ones were used. The alloys were vacuum-melted and hot-rolled to 3-mm thick plates. They were sliced and drawn into wires of 2-mm diameter, which were cut to 42 mm in length for torsion test.

2.2. Apparatus and Procedures

The schematic setup of X-ray diffraction experiment during torsional deformation is shown in Fig. 1. In this investigation, McScience MXC-18 with a cobalt anode was used as the X-ray source. The accelerating voltage and the current was 45 kV and 280 mA, respectively.

The wire specimen was heated from both sides by two 1 kW infrared heaters with quartz introducing rods of 20 mm diameter. A thermocouple was attached onto the center of the heated zone of the specimen, and the temperature of the specimen was controlled either automatically or manually. Specimens were coated with flux to reduce oxidation. Austenitization was performed for more than 3 min at 790–890°C, which were above the $A_{13}$ temperatures of the alloys. Specimens were then cooled to deformation temperatures at the rate of about 1°C/s and deformed in torsion. Some specimens were deformed at the same temperatures as the heating before deformation was performed.

A speed-controlled motor was used for torsional deformation. The plastic strain rate at the specimen surface ranged between 0.01 and 0.09 s⁻¹. Deformation time was 1 to 5 min.

The specimen was exposed to the X-ray beam usually during torsional deformation period. X-ray diffraction lines were recorded on a film in a flat camera shielded from heat with thin aluminum foils. Specimens in the cold state before and after deformation, and also those heated without deformation were also measured to confirm the identification of diffraction lines.

The torque acting to the specimen during torsional deformation was measured by a bending load cell connected to the specimen grip opposite to that being driven. Shear stress at the surface, $\tau$, calculated from the measured torque, and shear strain at the surface $\gamma$, estimated from rotation angle, were then converted into true stress $\sigma$ and true strain $\varepsilon$, respectively, by formulas $\sigma = \sqrt{3} \tau$ and $\varepsilon = \gamma / \sqrt{3}$, assuming the von Mises’ yielding criterion.

After deformation and heating had stopped, specimens were quenched by helium gas within 0.5 s. The cooling rate always exceeded 100°C/s. The deformed specimens were cut at the center of heating area. The microstructure was examined along the center plane in the longitudinal direction with optical microscope.

3. Results

3.1. In-situ X-ray Diffraction Experiment

Table 1. Chemical compositions of the alloys tested. (mass\%)

<table>
<thead>
<tr>
<th>Alloy</th>
<th>C</th>
<th>Ni</th>
<th>Si</th>
<th>Mn</th>
<th>P</th>
<th>S</th>
<th>Al</th>
</tr>
</thead>
<tbody>
<tr>
<td>A</td>
<td>0.0008</td>
<td>5.95</td>
<td>0.006</td>
<td>&lt;0.005</td>
<td>&lt;0.001</td>
<td>0.0002</td>
<td>0.015</td>
</tr>
<tr>
<td>B</td>
<td>0.096</td>
<td>5.98</td>
<td>0.008</td>
<td>&lt;0.005</td>
<td>&lt;0.001</td>
<td>0.0002</td>
<td>0.015</td>
</tr>
<tr>
<td>C</td>
<td>0.29</td>
<td>6.01</td>
<td>&lt;0.005</td>
<td>&lt;0.005</td>
<td>&lt;0.001</td>
<td>0.0002</td>
<td>0.015</td>
</tr>
</tbody>
</table>

Table 2. Transformation temperatures calculated and measured. (°C)

<table>
<thead>
<tr>
<th>Alloy</th>
<th>Calculated</th>
<th>Measured</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>$A_1$, $A_2$, $A_3$, $A_s$, $A_{15}$, $A_{13}$, $M_s$</td>
<td>$A_1$, $A_2$, $A_3$, $A_{15}$, $A_{13}$, $M_s$</td>
</tr>
<tr>
<td>A</td>
<td>753 697 697</td>
<td>772 640</td>
</tr>
<tr>
<td>B</td>
<td>734 680 644</td>
<td>747 630 397</td>
</tr>
<tr>
<td>C</td>
<td>700 654 544</td>
<td>712 590</td>
</tr>
</tbody>
</table>

Fig. 1. A schematic diagram of the in-situ X-ray experiment setup.

© 2000 ISIJ
There appeared only a line corresponding to (111)\text{g-Fe} in the diffraction pattern besides the lines indexed to oxides. When torsional deformation was applied at the same temperature, however, there appeared a line corresponding to (110)\text{a-Fe} together with the (111)\text{g-Fe} line as shown in Fig. 2(c). This clearly shows that \text{g-f} transformation occurred during hot deformation. The same results were obtained also at higher temperatures (Figs. 2(d) and 2(e)). In the latter, deformation temperature was above \(A_{p3}\).

Figure 3 summarizes the results of in-situ X-ray experiments. In Alloy A (110)\text{a-Fe} did not appear in the deformation above \(A_{p3}\). Increase in strain rates seems to widen the region of dynamic \(\gamma\rightarrow\alpha\) transformation to higher temperatures in Alloys B and C. In these alloys the transformation was observed to occur above \(A_{p3}\) when strain rate was the highest.

### Table 3. X-ray diffraction data of the specimen (Alloy B) during deformation at 680°C after cooling from 790°C.

<table>
<thead>
<tr>
<th>Distance from center spot (nm)</th>
<th>Observed lattice distance (nm)</th>
<th>Standard lattice distance (nm)</th>
<th>Error (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>18.0</td>
<td>0.372</td>
<td>0.360 \text{Fe}_2\text{O}_3</td>
<td>3.33</td>
</tr>
<tr>
<td>23.9</td>
<td>0.298</td>
<td>0.287 \text{Fe}_2\text{O}_3</td>
<td>3.83</td>
</tr>
<tr>
<td>27.4</td>
<td>0.270</td>
<td>0.270 \text{Fe}_2\text{O}_3</td>
<td>0.00</td>
</tr>
<tr>
<td>30.3</td>
<td>0.252</td>
<td>0.253 \text{Fe}_2\text{O}_3</td>
<td>-0.39</td>
</tr>
<tr>
<td>38.1</td>
<td>0.220</td>
<td>0.215 FeO</td>
<td>2.32</td>
</tr>
<tr>
<td>42.8</td>
<td>0.206</td>
<td>0.208 (111)\text{g-Fe}</td>
<td>-0.96</td>
</tr>
<tr>
<td>44.9</td>
<td>0.202</td>
<td>0.203 (110)\text{g-Fe}</td>
<td>-0.49</td>
</tr>
</tbody>
</table>

Note: Camera length=34.3 mm, \(\lambda=0.17900\) nm.

3.2. Deformation Stress and Microstructure

Figure 4 shows examples of true stress vs. true strain curves corresponding to the above X-ray experiments in Alloy B. In this figure austenitizing temperature was 790°C.
for solid lines, and it was the same as deformation temperature for dotted lines. All the curves exhibit similar forms. The deformation stress nearly reaches its maximum value at 0.2 to 0.4 strain and then remains at the same level until larger strain. Above 720°C the deformation stress increases with decreasing temperature, as usually expected, while below 720°C it does not increase as rapidly as expected from the tendency at higher temperatures.

The situation will be seen more clearly in Fig. 5, where the deformation stress at 0.5 strain was plotted against deformation temperature. All the curves exhibit similar forms. The deformation stress nearly reaches its maximum value at 0.2 to 0.4 strain and then remains at the same level until larger strain. Above 720°C the deformation stress increases with decreasing temperature, as usually expected, while below 720°C it does not increase as rapidly as expected from the tendency at higher temperatures.

The situation will be seen more clearly in Fig. 5, where the deformation stress at 0.5 strain was plotted against deformation temperature. All the curves exhibit similar forms. The deformation stress nearly reaches its maximum value at 0.2 to 0.4 strain and then remains at the same level until larger strain. Above 720°C the deformation stress increases with decreasing temperature, as usually expected, while below 720°C it does not increase as rapidly as expected from the tendency at higher temperatures.

The deformation stress nearly reaches its maximum value at 0.2 to 0.4 strain and then remains at the same level until larger strain. Above 720°C the deformation stress increases with decreasing temperature, as usually expected, while below 720°C it does not increase as rapidly as expected from the tendency at higher temperatures.

The situation will be seen more clearly in Fig. 5, where the deformation stress at 0.5 strain was plotted against deformation temperature. All the curves exhibit similar forms. The deformation stress nearly reaches its maximum value at 0.2 to 0.4 strain and then remains at the same level until larger strain. Above 720°C the deformation stress increases with decreasing temperature, as usually expected, while below 720°C it does not increase as rapidly as expected from the tendency at higher temperatures.

For solid lines, and it was the same as deformation temperature for dotted lines. All the curves exhibit similar forms. The deformation stress nearly reaches its maximum value at 0.2 to 0.4 strain and then remains at the same level until larger strain. Above 720°C the deformation stress increases with decreasing temperature, as usually expected, while below 720°C it does not increase as rapidly as expected from the tendency at higher temperatures.

The situation will be seen more clearly in Fig. 5, where the deformation stress at 0.5 strain was plotted against deformation temperature. All the curves exhibit similar forms. The deformation stress nearly reaches its maximum value at 0.2 to 0.4 strain and then remains at the same level until larger strain. Above 720°C the deformation stress increases with decreasing temperature, as usually expected, while below 720°C it does not increase as rapidly as expected from the tendency at higher temperatures.

Fig. 5. Temperature dependence of true deformation stress of Alloy B at 0.5 strain. Solid lines represent deformation stress after austenitization. Austenitizing temperature was 790°C except for the points with asterisks for which austenitizing temperature was the same as deformation temperature. Thick broken lines represent deformation stress in the experiments without austenitization.

Fig. 6. Optical micrographs of the quenched specimens of Alloy B. Austenitized at 790°C and cooled to and held at 640°C (a), cooled to and deformed at 640°C (b), and cooled to and deformed at 720°C ((c), (d)). Shear strain rate was 0.01 s⁻¹ for (b) and (c), and 0.09 s⁻¹ for (d).
honing at 640°C. When deformation is applied to the specimen, however, a large amount of a small equi-axed constituent appears up to 720°C (Figs. 6 (b), 6(c), 6(d)). This constituent must be ferrite judging from the results of the above in-situ X-ray experiments. Especially at higher deformation temperatures, most part of the ferrite is as fine as 1 to 3 μm as in Figs. 6 (c) and 6(d).

4. Discussion

4.1. Evidence of DT and Abnormality in Deformation Stress

The appearance of the (110)\textsubscript{γ} Fe X-ray diffraction line during hot deformation of austenite clearly showed transformation of γ into α. Abnormality in deformation stress corresponded to the X-ray results. In order to analyze this abnormality, the deformation stress of single phase austenite below 720°C will be estimated by extrapolating the temperature dependence of deformation stress of austenite between 720 and 830°C.

We will assume the following conventional formula, which usually was successfully used for expressing the temperature and strain-rate dependence of hot deformation

\[ \sigma = A (k \exp(E/RT))^\alpha, \]

where constants A and p are constants, which were determined from experimental data, E is the activation energy and R the gas constant. 284 kJ/mol was taken as the value of E.\textsuperscript{14} This formula was shown to hold as far as dynamic recrystallization did not contribute to deformation stress.\textsuperscript{15} Therefore calculation was made for the conditions of the present study using a computer program that predicts microstructure and deformation stress of low carbon austenite.\textsuperscript{16} The results showed that in the present study dynamic recrystallization would not contribute to deformation stress substantially. The predicted values of deformation stress are shown as thin broken lines in Fig. 5.

Thick broken lines in Fig. 5 lie far beneath the above lines and also the measured values. In these experiments, the microstructure was confirmed to be completely ferritic at 640°C and mostly ferritic at 680 and 720°C from microstructure. The measured values of deformation stress below 720°C lie between the thick and thin broken lines, and become nearer to the former with decreasing temperature. Thus the deviation of deformation stress from the thin broken lines (single phase austenite) can be explained as the result of transformation of γ into α, which is much softer than γ, in consistence with the results of in-situ X-ray diffraction experiments.

The microstructure shown in Figs. 6(b) to 6(d) also supports the above features in deformation stress. Although there will be some growth of ferrite grains during quenching, it is very difficult to think such a large amount of equi-axed ferrite was formed during quenching. This ferrite was not apparently observed in the isothermal holding without deformation at the same temperature. Therefore most part of the ferrite must have been formed during deformation.

4.2. Effect of Strain Rate

In Fig. 3, the diffraction line corresponding to (100)\textsubscript{α,Fe} appeared only at the highest strain rate above \( A_p^3 \) in Alloys B and C. The deformation stress (Fig. 5) and the microstructure (Fig. 6), however, did not exhibit remarkable dependence on strain rate. Considering that the X-ray source used in this study was not strong enough as to always allow for weak diffracted beams to transmit specimens, we can say at least that the increase in strain rate is not disadvantageous to DT. This leads us to the reasoning that long-range diffusion of carbon, which usually accompanies \( \gamma \rightarrow \alpha \) transformation below \( A_p^3 \) as predicted by the Fe-C phase diagram, will not essentially be involved in DT. As discussed earlier, the formation of ultra-fine ferrite was found to be insensitive to the strain rate up to 250 s\textsuperscript{-1} in the preceding studies.\textsuperscript{21} Carbon diffusion should have been essentially suppressed in such high strain-rate deformation. As mentioned above, in this experiment also, higher strain rate showed a beneficial effect on DT. Therefore carbon diffusion would not play an important role in DT. Of course the diffusion of nickel, which is much slower than that of carbon, should also be neglected, so we can also discard \( A_p^3 \), which is calculated by assuming the partition of all the elements constituting the system, as the related transformation temperature.

In iron alloys two kinds of transformation that do not accompany long-range diffusion of carbon are known. One is so-called massive ferrite transformation, which is known to occur just below \( T_B \). This transformation, however, is observed only in iron alloys with very low carbon content, e.g. as in Alloy A in this study, but not in alloys containing carbon more than the maximum solubility limit of carbon in α-iron (0.02 mass%) in this alloy system.

The other is martensite transformation. Its starting temperature, \( M_s \), is known to be 200–300°C below \( T_B \). This transformation, however, is observed only in iron alloys with very low carbon content, e.g. as in Alloy B \( M_s \) is 397°C, which is 247°C lower than its \( T_B \) as shown in Table 2). Although \( M_s \) is known to rise somewhat by plastic deformation, the observed rise is usually within some 80°C.\textsuperscript{17} It is apparent that the present results can not be explained from either of these transformations at least in Alloys B and C.

4.3. Driving Force of DT

As shown in Fig. 3, DT was shown to occur even above \( A_p^3 \) in Alloys B and C. Furthermore, thermodynamical transformation temperatures in the DT should be \( T_B \) as discussed above, which are considerably lower than \( A_p^3 \) in these alloys. This means that in many of the present experiments DT occurs when chemical driving-force had negative values.
Therefore, energy must have been supplied from other source in these cases. In the present study, one of the possible sources is the mechanical work applied to the specimens during deformation. The first law of thermodynamics assures us that mechanical work can increase internal energy, and also free energy of the system. Energy of mechanical work, therefore, will be calculated in some typical cases and compared with the chemical driving force required to cause the DT.

Energy of plastic work, given by the area A under the measured stress-strain curve as demonstrated in Fig. 7, was calculated and its values are listed in the first column (A) of Table 4. Here the transformation start strain $\varepsilon_s$ and deformation finish strain $\varepsilon_f$ were taken as 0.2 and 1.2, respectively.

The free energy required to cause DT can be obtained as the sum of the deficient chemical free energy and the energy dissipated at the interface. The energy dissipated at advancing transformation interfaces, $G^B$, the amount of free energy dissipated at the interface, is given as

$$G^B = \nu \cdot V_m / M, \quad \text{(2)}$$

where $\nu$ is the migration rate of interface, $V_m$ the molar volume and $M$ the nobility of interface.\(^{[8]}\)

The formula $M = 0.035 \cdot \exp(-17700/T)$, given by Hillert for iron alloys,\(^{[18]}\) was used for calculating $M$.

From the postulation (i) $G^B$, the amount of free energy dissipated at the interface, is given as

$$G^B = e^{\frac{\Delta G^{+\alpha}}{RT}} \cdot \frac{\nu}{\varepsilon}, \quad \text{(3)}$$

where $X$ is the fraction transformed during time $t$ ($=\varepsilon/\varepsilon$) and $S$ the nucleation site which can be roughly estimated by $6d_y^2$ from the austenite grain size $d_y$, when deformation is not so large. Here $d_y$ was taken as 40 $\mu$m from microstructural observation. $X$ was estimated from deformation stress shown in Fig. 6, by postulating that the value of deformation stress at 640°C without austenitization is that of 100% ferrite and deformation stress increases linearly with increasing fraction of austenite. The values of $X$ obtained are included in Table 4. The values of $\nu$ calculated from (2) by using the above quantities are listed in Table 4. The values of $\Delta G^B$ calculated from (1) are also listed in Table 4 (C).

They are very small compared with (B).

The energy required to cause DT, $G_{req}$, is given as $X \cdot (B+C)$ and its values are listed in Table 4 (D). They are only 3.8 to 5.9% of the values of deformation energy (A). These results do not contradict with the common understanding that over 90% of energy of plastic work are converted into heat.

The amount of mechanical energy saved by DT, $G_{savr}$, is given by the area B between the deformation stress curve of supposed austenite (calculated from (1)) and that observed, as in Fig. 7. Calculated values are shown in Table 4 (E). $G_{req}$ (D) is only 11 to 14% of $G_{savr}$ (E). Thus it was shown that total energy of deformed material was greatly saved by DT even when chemical-driving force was negative (above $T_0$). The conclusion of the above analysis strongly indicates that the DT occurs in order to reduce total energy of deformed material, as dynamic recovery and dynamic recrystallization do.

### 4.4. Effect of Carbon Content

In Fig. 3, it is seen that in Alloy A the dependency of strain rate on ferrite formation is reverse to that in Alloys B and C. The reason for this might be possible contribution of massive ferrite transformation. In this alloy, $A^p_{1}$ is nearly equal to $T_0$, so the transformation can start without being induced by deformation (addition of deformation energy). Because the rate of massive transformation is usually reported to be very high, it may be able to start and proceed to a considerable degree before DT starts at slower strain rates. It can be also accelerated by hot deformation. In the present X-ray experiment, it is difficult to know how the amount of the transformed product increases with time due to insufficient power of the source. A much stronger X-ray source, such as one from synchrotron orbital resonance (SOR), will be able to clarify the kinetics of DT and related transformations.

Also only in Alloy A, $\gamma \rightarrow \alpha$ transformation was not observed above $A^p_{1}$ by the in-situ X-ray experiment. The reason for this is not clear, but it may be pointed out that in this alloy the $A^p_{1}$ temperature is higher than those of the other alloys are and therefore the deformation temperature just above $A^p_{1}$ is higher than those in the other alloys are. This corresponds to lower deformation stress at the deformation temperatures, so that deformation stress at the temperature just above $A^p_{1}$ in Alloy A is lower than that in the other two alloys.

The present authors had proposed that a critical stress level might exist for starting DT.\(^{[20]}\) It may be that a certain amount of dislocation accumulation is necessary to start DT. Further study in various alloy systems will be needed to prove this reasoning.

<table>
<thead>
<tr>
<th>Deformation temp. / °C</th>
<th>Energy of pl. work (A)</th>
<th>$\Delta G^{+\alpha}$ (B)</th>
<th>$X$</th>
<th>$\nu$ / 10$^4$ m/s</th>
<th>$G^B$ (C)</th>
<th>$G_{req}$ (D)</th>
<th>$D/A$ / %</th>
<th>$G_{savr}$ (E)</th>
<th>$D/E$ / %</th>
</tr>
</thead>
<tbody>
<tr>
<td>640</td>
<td>1473</td>
<td>-23</td>
<td>0.78</td>
<td>4.5</td>
<td>0.024</td>
<td>-18</td>
<td>720</td>
<td></td>
<td></td>
</tr>
<tr>
<td>680</td>
<td>1482</td>
<td>124</td>
<td>0.45</td>
<td>1.8</td>
<td>0.004</td>
<td>56</td>
<td>3.8</td>
<td>509</td>
<td>11.0</td>
</tr>
<tr>
<td>720</td>
<td>1250</td>
<td>240</td>
<td>0.31</td>
<td>1.1</td>
<td>0.001</td>
<td>74</td>
<td>5.9</td>
<td>522</td>
<td>14.1</td>
</tr>
</tbody>
</table>
5. Conclusion

Direct evidence of dynamic $\gamma \rightarrow \alpha$ transformation during hot deformation was obtained by *in-situ* X-ray diffraction experiments in Fe–6mass%Ni–(0.0008–0.29)mass%C alloys above their cooling transformation temperatures. Microstructural study also indicated that a large amount of fine-grained ferrite was formed during transformation. The analysis of deformation stress and chemical driving force of the transformation showed that DT occurred in order to reduce the total energy of deformed material since the deformation of energy of $\alpha$ was shown to be considerably smaller than that of $\gamma$, and the amount of deformation energy saved by the transformation was calculated to be much greater than the chemical energy consumed by the transformation at the tested temperatures.

Acknowledgments

We thank Messrs. K. Tanaka, S Adachi and S. Fukumuro for their assistance to the experiments, and Drs. T. Senuma and M. Suehiro for supplying materials and calculating transformation temperatures and free energy change by Thermo-Calc. The Ministry of Education of Japan, the Japan Society for the Promotion of Science and Nippon Steel Corporation supported the work.

REFERENCES


