A Novel Observation of Strain-Induced Ferrite-to-Austenite Retransformation after Intercritical Deformation of C-Mn Steel

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It was found that a ferrite-to-austenite retransformation can take place during isothermal holding after deformation in 0.19 pct C-1.46 pct Mn steel at higher intercritical temperatures, but it was not observed at lower intercritical temperatures. This new finding was revealed by the combined techniques of optical observation and quantitative metallographic measurements. Also, two unexpected experimental observations on the quenched microstructure can be interpreted very well by the retransformation: one is that microhardness does not decrease at the end of relaxation as expected; the other is that an experimentally observed X-ray peak gets broader after long relaxation. In contrast, dilatometry measurements on the undeformed specimens show that the thermodynamically induced transformation during the isothermal holding should be from austenite to ferrite. Therefore, the ferrite-to-austenite retransformation after deformation should be strain induced. Such a retransformation can be explained quite well using a validated and quantitative model for the intercritical rolling and softening of both austenite and ferrite. The model presented is based on the fact that in intercritically deformed steel there is a strain distribution over the two phases with the more strain in the softer ferrite. The ferrite dynamically recovers, which leads to a shift in the relative amount of stored energy in the two phases. This phenomenon leads to a shift in the equilibrium fractions of ferrite and austenite.

I. INTRODUCTION

IT is well known that plastic deformation applied to C-Mn steels in the austenitic region enhances the austenite-to-ferrite transformation since plastic deformation introduces more potent nucleation sites and leads to an increased driving force for transformation.^[1] Such an acceleration in the transformation kinetics is usually observed in cooling experiments after austenitic deformation with the following characteristics: both the transformation starting and finishing temperatures increase,^[2,3] the temperature range of the transformation contracts,^[3] the final fraction of polygonal ferrite increases,^[4] and the final ferrite grain size decreases.^[2,3] Yada *et al.*^[5,6] even found that dynamic austenite-to-ferrite transformation could occur during hot torsion of Fe-Mn-C alloys at temperatures above Ar₃. Besides undergoing transformation, the severely deformed material can also recrystallize. As transformation is a similar process to recrystallization in consuming the strain energy,^[7] they may compete for stored energy during the process of softening. That is, the softening process of the deformed material might be related to the deformation-induced transformation.

In order to investigate the softening kinetics after deformation, the stress relaxation technique has been frequently used due to the fact that it is an accurate *in-situ* method for studying the process of softening.^[8,9] This technique is capable of separating the contributions to softening arising from recovery and recrystallization. In this investigation, besides stress relaxation experiments, quantitative metallography, X-ray diffraction, microindentation, and electron backscattering diffraction (EBSD) were also used to separate the softening behavior of each of the two phases and the phase transformation more clearly. Furthermore, dilatometry measurements have been done to investigate the ferrite/austenite transformation during isothermal holding in the undeformed specimen. Using the combined techniques, a novel observation of ferrite-to-austenite transformation, which takes place during the isothermal holding after the intercritical deformation and is termed as straininduced retransformation, was revealed. This observation was analyzed using the Bergström dislocation model, the strain distribution between the two phases, and phase equilibrium calculations using ThermoCalc.

II. EXPERIMENTAL

The composition of the C-Mn steel studied in this work was 0.19wt pct C-0.445wt pct Si-1.46wt pct Mn-0.033wt pct Al. Its transformation temperatures of A_1 and A_3 were calculated by Thermo-Calc* as 705 °C and 809 °C, respectively.

*Thermo-Calc Software Inc. is in SE-113 47 Stockholm, SWEDEN.

Cylindrical specimens of length 10 mm and diameter 12 mm were used for the relaxation tests on a Gleeble 3500** ther-

**GLEEBLE is a trademark of DSI, Inc., Poestenskill, NY.

momechanical simulator. Specimens were first austenitized at 1100 °C for 3 minutes, and then cooled to 679 °C, at which, according to the dilatometry measurements, almost 50 pct ferrite should be present. This is to ensure that a similar morphology is produced at the different deformation temperatures. After a 10-minute isothermal hold at this temperature to attain equilibrium, each sample is brought to the deformation temperatures ranging from 525 °C to 900 °C and held for 5 minutes before deformation. ISO-T anvils were used to ensure that the temperature difference along the 12-mm length of each

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sample was within 10 °C after deformation.^[10] All specimens were deformed in compression to $\varepsilon = 0.5$ at a strain rate of 0.1 s⁻¹. Subsequent to the deformation, the samples were maintained at the deformation temperature and subjected to an extended isothermal hold in order to monitor the stress relaxation response. Finally, the specimens were quenched by water jets after isothermal holding for various times after the deformation. The maximum isothermal holding time was 50 minutes. In this investigation, specific attention was paid to the microstructures on the diametric section from the middle of samples quenched from the intercritical temperatures of 775 °C and 725 °C. First, they were observed under the optical microscope after standard polishing and etching. Then, quantitative metallography was performed in order to determine the phase fractions after etching by 5 pct nital. For each specimen investigated, the ferrite fractions were measured from at least ten photos by using the image processing software of Leica Qwin (the product of Leica Imaging System Ltd., Cambridge, England). Next, microhardness values were obtained for the ferrite phases in the down-quenched samples using microindentation tests under a load of 50 g. A minimum of 50 measurements were made in each sample. Also, X-ray diffraction using cobalt radiation was performed in some selected quenched specimens in order to follow the evolution of X-ray line broadening, which varies with the constituents of the microstructure, *i.e.*, martensite and ferrite, and dislocation density. Next, the EBSD technique was applied to make orientation image mapping in the quenched microstructures in order to determine the misorientation between grains or subgrains.

To separate the effects due to deformation from the "natural" austenite/ferrite transformation, *i.e.*, transformation behavior for an undeformed sample, we performed dilatometer tests imposing the same temperature profile but not imposing a deformation. To this aim, we used the Bähr 805 (BÄHR Thermoanalyse GmbH, Altendorfstaße 12, D-32609 Hüllhorst, Germany) dilatometer and 10-mm-long cylindrical specimens with a diameter of 6 mm.

III. RESULTS

A. Stress-Strain and Stress Relaxation Curves

At temperatures of 725 °C and 775 °C, the microstructure is comprised of 42 pct ferrite + 58 pct austenite and 25 pct ferrite + 75 pct austenite, respectively, by the measurement on the samples immediately quenched after deformation. During the deformation, both stress-strain curves suggest that the material does not recrystallize since the stress does not decrease (Figure 1(a)). In contrast, the relaxation curves after deformation are different (Figure 1(b)). At 775 °C, the stress-log(time) curve exhibits three distinct regions. The stress decrease immediately after deformation is linear as a function of log(time) and this linear softening continues for up to around 1 sec. It is widely accepted that such linear softening is due to recovery.^[8,9] Subsequently, there is acceleration in the stress decrease, which may be attributed to recrystallization, as proposed by Karjalainen et al.^[8,9] Finally, toward the end of the relaxation, there is a second inflection and the stress-log(time) curves tend to become less steep and once again linear. The second inflection is usually taken to signify the completion of recrystallization. On the other hand, at 725 °C, no inflection in the early stages of relaxation is observed and the curves comprise two linear parts with the change in slope occurring around 100 seconds. This observa-



Fig. 1—(*a*) Stress-strain curves during the intercritical deformation and (*b*) stress-log(time) curves during the relaxation after the deformation at temperatures of 725 $^{\circ}$ C and 775 $^{\circ}$ C.

tion suggests that austenitic and ferritic recovery both start after deformation. The change in slope of the two linear parts is explained by nonsynchronization of recovery in two phases: austenitic recovery finishes in a few minutes while ferritic recovery lasts a much longer time.^[11]

B. Optical Observation

Optical observations provided the evidence that austenite does recrystallize during the relaxation at 775 °C since some new small grains were found at around 4 seconds after deformation in the austenite phase (Figures 2(a) and (b)). Furthermore, it was frequently observed that some dark martensite islands as small as just a few micrometers appeared in the previously continuous and blocky ferrite phase after 3000 seconds of relaxation (Figures 2(c) and (d). This is easily recognized as an obvious difference with the microstructure in the specimen immediately quenched after deformation. By following the similar heat treatment, evolution of proeutectoid ferrite microstructure at 775 °C has been examined in the case of no deformation applied, as shown in Figures 2(e) and (f). When the pro-eutectoid ferrite morphology in specimen, quenched after 55 minutes isothermal holding at 775 °C after heating from 679 °C, is compared with



Fig. 2—Microstructures in the quenched specimens: (*a*) and (*c*) immediately quenched after deformation at 775 °C, (*b*) relaxed for 4 s and (*d*) relaxed for 3000 s and then quenched after deformation at 775 °C, (*e*) quenched after isothermal holding at 775 °C for 5 min without deformation, and (*f*) quenched after isothermal holding for 55 min without deformation. (a) and (b) Etched by saturated picric acid. (c) through (f) Etched by 5 pct nital.

that quenched after 5 minutes isothermal holding, such kind of morphology with dark islands in white phase is seldom observed. Finally, metallographic quantitative measurements on phase fractions are also consistent with this observation: the ferrite phase fraction starts to decrease after the relaxation at 775 °C for several minutes while it does not at 725 °C, as shown in Figure 3. Therefore, optical observations suggest the occurrence of ferrite-to-austenite retransformation too.

C. Microhardness and X-Ray Diffraction Measurements

Microhardness measurements, which were taken on the ferrite phase in the specimens quenched from 725 °C and 775 °C after the various hold times of stress relaxation, are shown in Figure 4(a). The ferritic hardness decreases continuously during the relaxation at 725 °C, which is consistent with the ferritic recovery. However, the ferrite hardness does not decrease as expected at the end of the relaxation at 775 °C. Furthermore, the X-ray peak of {220} reflection from the mixture of marten-



Fig. 3—The ferrite fraction during the relaxation at temperatures of 775 $^{\circ}$ C and 725 $^{\circ}$ C, measured by quantitative metallography.



Fig. 4—(*a*) Analysis on the quenched specimens by microindentation and X-ray diffraction. The microhardness during the relaxation at 725 °C and 775 °C. (*b*) X-ray line broadening of the $\{220\}$ reflection after the relaxation for 3000 s.



Fig. 5—EBSD results from the specimen (*a*) and (*b*) immediately quenched after deformation at 775 °C and (*c*) and (*d*) quenched at 50 min isothermal holding after deformation at 775 °C. (a) Secondary electronic image; (b) EBSD boundary misorientation image after mask in the austenite (martensite) phase, in which misorientation levels of 0.55, 1, and 5 deg correspond to thinnest, medium thick, and thickest lines in the map; and (c) EBSD band contrast image to reveal the ferrite and martensite EBSD misorientation mapping image, in which misorientation levels of 1 and 15 deg correspond to thin and thick lines in the map.

site and ferrite becomes slightly broader after relaxation for 3000 seconds at 775 °C, as shown in Figure 4(b). This seems to suggest that more dislocations are formed or constituent of microstructure, *i.e.*, martensite and ferrite, changes when the

specimen is isothermally held for 3000 seconds and then quenched, which will be discussed in Section A of Discussion. Both observations are difficult to understand because recovery and recrystallization, taking place during the relaxation after the deformation, should lead to a reduced dislocation density and consequently a continuously decreasing hardness in the microstructure and smaller X-ray peak width.

D. EBSD Results

Orientation image mapping in both phases was performed on the specimens immediately quenched or relaxed for 50 minutes and then quenched after deformation at 775 °C, and the results are shown in Figure 5. This misorientation boundary mapping gives an approximate grain structure in ferrite but not in martensite, which was transformed from austenite. Measurements on misorientation in the ferrite microstructure with an area of around $15 \times 10 \ \mu m^2$ show that ferrite within this area only comprises low-angle boundaries with the misorientation less than 15 deg (Figure 5(b)), *i.e.*, only subgrains with the size of a few micrometers were formed in ferrite after the intercritical deformation, even though they sometimes look like grains under the optical microscope after etching. This suggests that ferrite just dynamically recovers during the deformation. The band contrast image taken from the specimen quenched after 50 minutes relaxation also clearly reveals the martensite islands appearing within the ferrite phase, as shown in Figure 5(c). Misorientation mapping images show that highangle boundaries are actually phase boundaries between ferrite and martensite while most of the boundaries within the continuous ferrite phase are still low angle, which suggests that ferrite statically recovers during relaxation.

E. Dilatometry Measurement

The temperature-time and dilation-time curves recorded in the dilatometer are shown in Figure 6. Concentrating on the final intercritical holding at 775 °C, we noted that the dilatation decreases at the beginning, *i.e.*, the ferrite fraction decreases due to the ferrite-to-austenite transformation. After 380 seconds, which is close to the holding time prior to deformation, the ferrite fraction slowly increases again. This increase is due to the slow transition from the paraequilibrium state to the full equilibrium state. It is important to note that the ferrite fraction slowly increases during the isothermal holding for undeformed material, which is different from that in the deformed material, as shown in Section B of Results.



Fig. 6—Dilatometry measurements on the "natural" retransformation during isothermal holding at 775 $^{\circ}\mathrm{C}$ without deformation

A. Observations of Strain-Induced Ferrite-to-Austenite Retransformation

The dilatometry measurements during the isothermal holding on the heated specimens indicate that a slow austeniteto-ferrite transformation should happen. Meanwhile, optical observation and quantitative metallographic measurements on the guenched microstructures seem to indicate that a ferriteto-austenite retransformation has taken place during the relaxation after deformation at 775 °C, resulting in dark martensite islands in the white ferrite phase. Such retransformation is also indirectly suggested by microhardness and X-ray diffraction measurements on the quenched samples. The retransformed austenite grains of a diameter of 1 to 2 μ m are within the ferrite phase (Figure 2(d)), which is too small to be identified under the microscope for making microindents because its maximum resolution is only 400 times. Therefore, when microindents are made on the quenched sample in which the retransformation has taken place, it is very likely that some indents, which were expected to be made in ferrite, are actually in the mixture of martensite and ferrite as the indents are around 20 imes $20 \ \mu m^2$. This can explain why the ferritic microhardness measured at 775 °C after 3000 seconds shows a higher value than expected from the observed trend (Figure 4(a)). Moreover, such a retransformation leads to more austenite formed, and therefore more martensite can be formed during the quenching, which leads to a broader X-ray peak (Figure 4(a)) due to the tetragonality of the martensite lattice and the slight difference between martensite and ferrite in their lattice parameters. Merely one technique employed in investigation might be misleading; since several techniques are combined and employed here, and they all lead to the same conclusion, the possibility to give wrong conclusions is minimized. Therefore, it is concluded that the retransformation from ferrite to austenite does indeed take place during the isothermal holding after the intercritical deformation, and it is strain induced since such retransformation cannot be observed in the undeformed specimen.

B. Mechanism for the Ferrite-to-Austenite Retransformation

Why can such a retransformation take place during the isothermal holding at 775 °C but not at 725 °C? It was noticed that austenite could recrystallize after deformation at 775 °C but only recover at 725 °C, while ferrite recovers at both temperatures. Pandi and Yue^[12] also observed a similar phenomenon that the dynamic recrystallization is accompanied by a decrease of austenite in the volume fraction of ferrite during the intercritical deformation. They thought that deformation heat is responsible for the observed decrease in ferrite fraction. In our case, however, such a decrease occurred during the isothermal holding, and therefore, deformation heat cannot be the reason. It is well known that deformation introduces dislocations into materials and thus increases the stored energy as part of the driving force for transformation. When deformation is performed in the intercritical region, the two phases deform differently and thus acquire a difference in stored energy. After deformation, the two phases soften differently and thus the stored energy in two phases decreases by different rates, which changes the phase equilibrium and leads to ferrite-to-austenite retransformation.

The next issue is how to quantify the variation of stored energy in the deformed austenite and ferrite during relaxation at 775 °C. In order to calculate the variation of stored energy in both phases, the strain division during the intercritical deformation has to be determined first, as the softening kinetics of each phase depends on the strain level in the phase. Then, kinetics of ferritic recovery and austenitic recrystallization during the relaxation is determined to quantify the variation of stored energy in each phase. The process for quantifying the change in phase equilibrium during the relaxation at 775 °C can be summarized as follows.

- (a) The Bergström dislocation model together with rules of mixture was applied to the intercritical region to derive the strain distribution between austenite and ferrite.
- (b) The change of stored energy in each phase was calculated by the determined kinetics of austenitic recrystallization and ferritic recovery from stress-log(time) relaxation curves after deformation.
- (c) Shift of phase equilibrium under the varied stored energies in the two phases was calculated by ThermoCalc.

C. Modeling the Strain Division by Bergström Dislocation Model and Rules of Mixture

Since the hot strength of ferrite is lower than that of austenite at the same temperature, austenite and ferrite will deform differently during the intercritical deformation. The Bergström dislocation model together with rules of mixture is employed here to calculate the strain in each phase, since a direct measurement of the exact strains in both phases is not feasible. Bergström's model describes the development of the dislocation density during deformation,^[13,14] which can be used for modeling the work hardening and dynamic recovery in a metal during deformation.^[15,16] When the initial stress and dislocation density in the material before deformation are ignored, Bergström's model can be used to describe the stress-strain curve, $\sigma(\varepsilon)$, as follows:^[15]

$$\sigma(\varepsilon) = \left[\frac{\psi(T)}{\Omega(T)} \cdot (1 - e^{-\Omega(T) \cdot \varepsilon})\right]^{1/2}$$
[1]

where the temperature-dependent parameters of $\psi(T)$ and $\Omega(T)$ can be calculated by

$$\psi(T) = (\alpha \cdot G(T) \cdot \mathbf{b}(T))^2 \cdot U \qquad [2]$$

$$\Omega(T) = k \cdot \varepsilon^n \cdot \exp\left(-Q_\Omega/\mathbf{R}T\right)$$
[3]

where U is the immobilization parameter of dislocations, and is used as a fitting parameter since its exact value is not known. The apparent activation energy, Q_{Ω} , is 17.3 kJ/mol, as reported by Yoshie *et al.*^[17] The exponent *m* is -0.1 for austenite^[17] and -0.3 for ferrite.^[15,16] The constant α has a weak dependence of temperature^[18] but is in the order of 0.5.^[19] The Parameter *k* is dependent on grain size, and is here used as another model fitting parameter. The term G(T) is the shear modulus, and is calculated by the following equations from Reference 15:

$$G_{\gamma}^{T} = 1.45 \cdot 10^{5} - 137 \cdot T + 3.48 \cdot 10^{-2} \cdot T^{2}, \text{ in MPa,}$$

for austenite [4]
$$G^{T} = 2.07 \cdot 10^{4} + \frac{5.84 \cdot 10^{4}}{10^{4}}$$

$$G_{\alpha} = 2.07 \cdot 10^{-4} + \frac{1}{1 + \exp(-7.80 + 9 \cdot 10^{-3} \cdot T)},$$

in MPa, for ferrite [5]

where *T* is temperature in Kelvin. Finally, $\mathbf{b}(T)$ is the Burgers vector, which is calculated as a function of temperature from the lattice parameters of austenite and ferrite (in nanometers),^[15] according to

$$\mathbf{b}_{\gamma}(T) = \frac{\sqrt{2}}{2} \cdot (0.3620 \cdot (1 + 24.73 \cdot 10^{-6} \cdot (T - 1000)))$$

for austenite [6]

$$\mathbf{b}_{\alpha}(T) = \frac{\sqrt{3}}{2} \cdot (0.28863 \cdot (1 + 17.55 \cdot 10^{-6} \cdot (T - 800)))$$

for ferrite [7]

where the temperature is given in Kelvin. Before the Bergström model is applied to the intercritical region, it needs to be applied in the single phase of austenite and ferrite to fit the stress-strain curves, through which the model fitting parameters U, α , and k can be determined (Figure 7). Since about 20 pct pearlite



Fig. 7—Stress-strain curves in the single phase of austenite and ferrite fitted by Bergström model: (*a*) in the austenitic region of the current investigated steel and (*b*) in the ferritic region of the ultra-low-carbon steel with the compositions of 0.052C-0.23Mn (wt pct) and experimental data from Ref. 15.

is formed in the ferritic region of the investigated steel and makes a considerable contribution to the stress during the deformation, data on the hot deformation behavior of ultra-low-carbon steel in Reference 15, which are supposed to be close to the intercritical ferrite, are chosen for this simulation. It needs to be mentioned that the parameter α is on the order of 0.5, but slightly varies with the temperature and has a significant influence on the calculated stress by the Bergström model. Roberts and Bergström^[18] proposed that α should increase as the temperature decreases. The term α , determined by least-squares fitting in the single-phase region, has an approximately linear dependence on temperature, as shown in Figure 8. This linear relationship will be extrapolated into the intercritical region. Parameters for the Bergström modeling calculation are summarized in the Table I. Finally, this model is further incorporated into the rules of mixture for the intercritical region given by

$$\begin{cases} f_{\gamma} \cdot \varepsilon_{\gamma} + f_{\alpha} \cdot \varepsilon_{\alpha} = \overline{\varepsilon} \\ f_{\gamma} \cdot \sigma^{\gamma} (\varepsilon_{\gamma}) + f_{\alpha} \cdot \sigma^{\alpha} (\varepsilon_{\alpha}) = \overline{\sigma} \end{cases}$$

$$[8]$$

where f_{γ} and f_{α} are ferritic and austenitic fraction, respectively; ε_{γ} and ε_{α} are the strain during the deformation in the unit volume of austenite and ferrite, respectively; and $\overline{\varepsilon}$ and $\overline{\sigma}$ are the bulk strain and stress in the sample, which can be measured and automatically recorded during the deformation. The strain in austenite or ferrite is then calculated. Calculations performed at 775 °C and 725 °C are shown in Figure 9 as an example. It is obvious that most strain is concentrated in the ferrite phase



Fig. 8—Constant α in the Bergström model shows linear dependence on the temperature in both the austenitic and the ferritic regions.

Table I. Model Parameters for the Bergström Model*

Parameter	Ferrite	ite Austenite	
U	$6.81 \cdot 10^{15}$	$2.87 \cdot 10^{16}$	m^{-2}
Q_{Ω}	17.3 ^[17]	17.3 ^[17]	kJ/mol
m	$-0.33^{[15,16]}$	$-0.1^{[17]}$	_
α	$-0.001105 \cdot T(K)$	$-0.000628 \cdot T(K)$	
	+ 1.5987	+ 1.0693	N/m ²
k	45.44	48.38	S
*The terms	s U, α , and k were deriv	ed by the least-squares	fittings of

stress-strain curves in the austenitic and ferritic single phase.

because ferrite is quickly dynamically recovered and reaches a low saturated stress level, while austenite is still in the state of work hardening. It appears that a smaller ferrite fraction, resulting from the higher intercritical temperature, and larger strain will both lead to more strain being located in ferrite.

D. Calculation of the Varied Stored Energies in Both Phases

Figure 10 explains how to calculate softening kinetics of austenite and ferrite at 775 °C. First, the stress of austenite and ferrite just after the deformation, *i.e.*, σ_{α}^{0} and σ_{γ}^{0} in Figure 10, can be calculated, respectively, by the Bergström model together with the rule of mixtures, as stated previously. Then, the stored energy, *E*, can be approximately calculated from the stress by the equation^[1,19]

$$E = C_2 \rho G \mathbf{b}^2 + \frac{3\gamma_s}{D}$$
[9]

where G and **b** are the shear modulus and Burger vector, and are calculated from Eqs. [4] and [5] and Eqs. [6] and [7], respectively;



Fig. 9—Strain distribution between austenite and ferrite during the intercritical deformation at (a) 775 °C and (b) 725 °C, calculated by the Bergström model and the rule of mixtures.



Fig. 10—Calculation on kinetics of ferritic recovery and austenitic recrystallization during the relaxation after deformation at 775 $^{\circ}\mathrm{C}.$

 γ_s is the interfacial energy of the subgrain with the value of 0.2 J/m².^[1] The term C_2 is a constant with the typical value of 0.5,^[19] and *D* is the diameter of the subgrains. For ferrite, subgrains do not grow so much during the relaxation (the optical observation shown in Figure 4), and thus, the subgrain size was fixed at 4 μ m for the calculations. For austenite, it can be reasonably assumed that no subgrains are formed since austenite was just work hardened during the deformation. Therefore, its contribution to stored energy by formation of austenitic subgrains can be ignored ($D \rightarrow \infty$). The term ρ is the dislocation density and can be calculated from the stress by the equation

$$\rho = \left(\frac{\sigma}{\alpha G \mathbf{b}}\right)^2 \tag{10}$$

where α is linearly dependent on the temperature and can be calculated as shown in Figure 9. Then, the expression for ρ in Eq. [10] can be inserted into Eq. [9], which yields

$$E = \frac{C_2}{\alpha^2} \frac{\sigma^2}{G} + \frac{3\gamma_s}{D}$$
[11]

It is known by our experimental observations that austenite recrystallizes and ferrite recovers after deformation at 775 °C. The ferritic recovery during the relaxation should also observe the empirical exponential relationship $\sigma = a - b \log (t)$ (Figure 11), where *a* and *b* are constants independent of time. If we assume that the stress at the end of relaxation, *i.e.*, σ_F , only results from the ferrite since austenite has completely recrystallized and just makes little contribution to the bulk stress, the constants *a* and *b* can be derived by solving

$$\begin{cases} \sigma_{\alpha}^{0} = a_{\alpha} - b_{\alpha} \log (t_{0}) \\ \sigma_{F} f_{\alpha}^{F} = a_{\alpha} - b_{\alpha} \log (t_{F}) \end{cases}$$
[12]

where t_0 and t_F are the time for starting and finishing points of relaxation, and are 0.1 and 3000 seconds, respectively, and f_{α}^F is the ferrite fraction at the end of relaxation. Actually, a_{α} and b_{α} at the starting stage of relaxation are a bit different from those at the finishing stage of relaxation since phase fraction changes a bit during the relaxation at 775 °C, as shown in Figure 3. In order to solve this issue, the average values of phase fractions at the starting and finishing points were employed in the calculation, and then one set of parameters a_{α} and b_{α}



Fig. 11—Calculation showing that variation of stored energy in austenite and ferrite leads to a reduced volume fraction of ferrite, and it is compared with experimental measurements.

Table II.The Parameters to Fit the Kinetics of Recovery
and Recrystallization at 775 °C

For Reco Stages by	overy at the ly the Linear	For Recrystallization by JMAK Equation			
$\overline{a_1}$	b_1	<i>a</i> ₂	b_2	п	<i>t</i> _{0.5} , s
141.9	31.5	70.5	16.6	0.73	12.3

could be derived for the entire process. Finally, the stored energy of ferrite during relaxation, E_{α} , is calculated by Eqs. [11] and [12] with the determined values of a_{α} and b_{α} . As austenite recrystallizes at 775 °C, the stored energy of austenite during the relaxation can be calculated by the following equation:

$$E_{\gamma} = E_0^{\gamma} (1 - f_{\text{Rex}})$$
 [13]

where E_0^{γ} is the initial stored energy of austenite just after deformation, which was calculated from the initial stress in a similar way as for ferrite. The term f_{Rex} is the recrystallized fraction of austenite, which can be expressed by the Johnson– Mehl–Avrami–Kolmogorov (JMAK) equation:

$$f_{\text{Rex}} = 1 - \exp\left(-0.69(t/t_{0.5})^n\right)$$
[14]

where *n* is the Avrami exponent and $t_{0.5}$ the time for accomplishment of 50 pct recrystallization. Meanwhile, the austenitic recrystallized fraction can be directly calculated from the stress-log(time) curve by using the approach described in Reference 8. These calculated recrystallized fractions were further fitted by the JMAK equation, by which *n* and $t_{0.5}$ in the JMAK equation could be determined (Figure 10). The parameters for the linear relationship in the first and last stages of stress-log(time) curve, resulting from static recovery, were derived by linear fitting and summarized together with *n* and $t_{0.5}$ in Table II. Next, the stored energy in austenite during the relaxation was calculated by Eq. [13].

E. Calculation of the New Phase Equilibrium by ThermoCalc

Finally, the equilibrium phase fractions, taking into account the stored energies, can be calculated by Thermo-Calc at several selected times and compared with measured values, as shown in Figure 11. The calculation results show the same tendency as the measurements, although there is a shift between them. Obviously, the calculation of the stored energy is an approximation since the constants in the model were just chosen with the typical values. Moreover, the new steel database with the title of "TC-Fe2000" in Thermo-Calc, which was used for this calculation, might not be a good fit for our specified steel. Finally, this difference could also result from the fact that the actual experimental conditions did not reach the full equilibrium, but are between full equilibrium and paraequilibrium because it takes a long time for substitutional elements to reach their equilibrium concentrations due to their slow diffusion. The decrease in the measured ferrite volume fraction shows a time delay to the calculated one. This is quite reasonable since the calculation is based on thermodynamics while measurement also reflects the kinetics of transformation.

In contrast, it was difficult to calculate the variation of stored energy in austenite and ferrite during the relaxation at 725 °C, because both phases just recovered and their kinetics were difficult to separate. If variation of stored energy in both phases at 725 °C was assumed to be the same as the one at 775 °C, calculations also showed that the stored energy had less influence on phase equilibrium at 725 °C than 775 °C (Figure 12). Moreover, thermodynamic calculations also show that full equilibrium only depends on the difference between the stored energy in two phases. Obviously, variation of this difference during relaxation at 725 °C should be smaller than that at 775 °C, because austenitic recrystallization at 775 °C should consume the stored energy more rapidly than austenitic recovery at 725 °C, leading to the maximum of variation in the difference between the stored energy at 725 °C smaller than that at 775 °C. In addition, transformation kinetics is slower at the lower temperature, leading to completion of such a small change in phase equilibrium taking much longer time, which might be more than the relaxation time. Both lead to a much smaller change in phase fraction during the relaxation at 725 °C, and therefore, it could not be observed by quantitative metallography, as shown in Figure 3.



Fig. 12—The difference between the stored energy in the two phases has less influence on phase equilibrium at 725 $^{\circ}$ C than at 775 $^{\circ}$ C.

V. CONCLUSIONS

Optical observation and quantitative metallography suggest the ferrite-to-austenite retransformation, *i.e.*, a decreasing ferrite fraction, during the relaxation after the intercritical deformation at 775 °C, but not at the lower temperature of 725 °C. This retransformation can also explain two unexpected experimental observations in the quenched microstructures: microhardness in ferrite does not decrease at the end of the relaxation and X-ray diffraction peak gets broader after long relaxation. In contrast to the deformed material, the undeformed material shows an increasing ferrite fraction over the same part of the intercritical annealing cycle. From this, we conclude that the observed ferrite-to-austenite retransformation after intercritical deformation is indeed strain induced.

Such a retransformation can be interpreted quite well by the modeling of intercritical rolling, the following softening kinetics, and changes in phase equilibrium. The modeling calculations suggest that variation of stored energy in both phases, resulting from the different softening kinetics in the two phases, is responsible for this retransformation. Since both the variation of differences between the stored energy in the two phases and its influence on phase equilibrium are smaller at lower temperatures, such a retransformation is not observed at 725 °C.

REFERENCES

- 1. D.N. Hanlon, J. Sietsma, and S. van der Zwaag: Iron Steel Inst. Jpn. Int., 2001, vol. 41, pp. 1028-36.
- X. Liu, L.P. Karjalainen, and J.S. Pertulla: Conf. Proc. 2nd Int. Conf. on Modelling of Metal Rolling Processes, J.H. Beynon, P. Ingham, H. Tecichert, and K. Waterson, eds., The Institute of Metals, London, 1996, pp. 489-500.
- R. Pandi, M. Militzer, E.B. Hawbolt, and T.R. Meadowcroft: Conf. Proc. 37th MWSP, ISS, Warrendale, PA, 1996, vol. XXXIII, pp. 635-47.
- A.J. De Ardo: Conf. Proc. Accelerated Cooling of Steels, P. Southwick, ed., AIME, Pittsburgh, PA, 1986, pp. 97-106.
- H. Yada, C.M. Li, and H. Yamagata: *Iron Steel Inst. Jpn. Int.*, 2000, vol. 40, pp. 200-06.
- C.M. Li, H. Yada, and H. Yamagata: Scripta Mater., 1998, vol. 39, pp. 963-67.
- H. Yada and T. Senuma: Proc. 7th JIM Int. Symp. on Aspects of High Temperature Deformation and Fracture Deformation and Fracture in Crystalline Materials, Nagoya; Japan; 28–31 July 1993, pp 51-58.
 Y. Hosoi, H. Yoshinaga, H. Oikawa, and K. Maruyama, eds., published by The Japan Institute of Metals, Aramaki Aoba, Aoba-Ku, Sendai 980, Japan.
- 8. L.P. Karjalainen: Mater. Sci. Technol., 1995, vol. 11, pp. 557-64.
- L.P. Karjalainen, J. Pettula, Y. Xu, and J. Niu: Proc. 7th Int. Symp. on Physical Simulation of Casting, Hot Rolling and Welding, Tsukuba; Japan, 1997, pp. 231-236, published by National Research Institute for Metals, 2-3-12, Nakameguro, Meguro-ku, Tokyo, Japan.
- "Axisymmetric Uniaxial Compression Testing Using ISO-T Anvils on Gleeble Systems," Application Note, DSI Company, Poestenkill, New York, NY, http://www.gleeble.com/AppNotes/AppNotes.
- H. Luo, D.N. Hanlon, J. Sietsma, and S. van der Zwaag: Proceedings of 4th Int. Conf. on Physical and Numerical Simulation of Materials Processing (CD-ROM), Shanghai, China, May 26–29, 2004.
- 12. R. Pandi and S. Yue: Iron Steel Inst. Jpn. Int., 1994, vol. 34, pp. 270-79.
- 13. Y. Bergström: Rev. Powder Metall. Phys. Ceram., 1983, vol. 2, pp. 79-265.
- 14. Y. Bergström: Mater. Sci. Eng., 1969-70, vol. 5, pp. 193-200.
- A. Bodin: Ph.D. Dissertation, Delft University of Technology, Delft, The Netherlands, 2002, pp. 19-23.
- A. Bodin, J. Sietsma, and S. van der Zwaag: Scripta Mater., 2001, vol. 45, pp. 875-82.
- A. Yoshie, H. Morikawa, Y. Onoe, and K. Itoh: *Trans. Iron Steel Inst. Jpn.*, 1987, vol. 27, pp. 425-31.
- 18. W. Roberts and Y. Bergström: Acta Metall., 1973, vol. 21, pp. 457-69.
- F.J. Humphreys and M. Hatherly: *Recrystallization and Related Annealing Phenomena*, Elsevier Science Ltd., Oxford, United Kingdom, 1995, pp. 16-17.