A Metallurgical Interpretation of the Static Recrystallization Kinetics of an Intercritically Deformed C-Mn Steel

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The austenite recrystallization kinetics in the intercritical region of a C-Mn steel is investigated by means of stress relaxation tests. It is found that the Avrami exponent, *n*, decreases significantly with decreasing temperature, *i.e.*, with increasing ferrite fraction. This behavior deviates from that of austenite recrystallization in the purely austenitic state, in which case the Avrami exponent is constant and independent of temperature. To resolve the origin of the changing Avrami exponent, the influence of the austenite/ferrite interface boundary area and that of the spatial variation of the plastic strain in the austenite grains is modeled quantitatively. The modeling results seem to indicate that the strain heterogeneity rather than the reduced interface boundary length is responsible for the decreasing Avrami exponent with decreasing temperature.

I. INTRODUCTION

MODERN high-quality steel strip production requires a detailed understanding of the softening kinetics during the interpass cycles in hot strip rolling to ensure accurate thickness and microstructure control. Hence, in the past, numerous studies^[1–5] have been devoted to the softening kinetics after deformation of fully austenitic microstructures as a function of temperature, strain, and strain rate. From these studies, it emerges that the recrystallization kinetics can be described by the Johnson-Mehl-Avrami-Kolmagorov (JMAK) model, which is one of the most frequently used classical models for the analysis of the kinetics of recrystallization and phase transformations.^[6] Furthermore, it turns out that the Avrami exponent for austenite recrystallization is close to 3 or 4 as JMAK theory predicts only in some lightly deformed fined-grained texture-free materials of a uniform grain size, most often varies between 1 and 2 for the various C-Mn steel grades, but shows no or only a slight dependence on temperature or deformation conditions.

In contrast, much less detailed knowledge exists on the softening kinetics during intercritical rolling, which is a relatively new rolling procedure applied in direct strip rolling operations or as a variant of classical thin hot-strip rolling. In intercritical rolling, the microstructure to be deformed consists of a mixture of ferrite and austenite. The starting structure therefore differs from that in austenitic rolling in that it now contains austenite-ferrite interface boundaries rather than just austenite grain boundaries. These austenite-ferrite grain boundaries may play a different role in nucleation of recrystallization than the austenitic grain boundaries, since strain gradients can be expected near the interface boundary.^[7,8] The dual-phase microstructure of the soft ferrite grains and the

harder austenite grains will lead to local plastic strain variations depending on the connectivity of the ferrite and austenite networks.^[9,10] Hence, the initial strain distribution in the austenite fraction just after deformation is intrinsically more inhomogeneous than in the case of deforming a fully austenitic structure.

The present work aims at investigating the softening kinetics of a C-Mn steel after deformation in the intercritical region in more detail using the so-called stress relaxation technique. The stress relaxation technique is an accurate *in-situ* method, which has been used frequently to investigate softening kinetics.^[11,12] It is capable of monitoring the softening process both at short and long time scales. The technique has been used successfully to separate the contributions to softening arising from ferrite recovery and those from austenite recrystallization.^[11,12,13]

In the present investigation, we concentrate on the recrystallization kinetics of the austenite fraction after deformation in the intercritical region. Surprisingly, it is found that the Avrami exponent n for intercritical austenite recrystallization kinetics is not constant, but a function of temperature and hence of phase fraction. In order to interpret this new phenomenon, we model both the influence of the austenite-ferrite interface boundary area and that of the strain heterogeneity in the austenite on the recrystallization kinetics and hence on the Avrami exponent for the macroscopic process.

II. EXPERIMENTAL

The composition of the C-Mn steel studied is 0.19 wt pct C-0.445 pct Si-1.46 pct Mn-0.033 pct Al, balance Fe. According to thermodynamic calculations using Thermocalc (Stockholm, Sweden) software and TC-Fe2000 database, its transformation points A_1 and A_3 are 705 °C and 809 °C, respectively. Cylindrical specimens of 10-mm length and 12-mm diameter were used for the relaxation tests on a Gleeble DSI, Inc., Poestenskill, NY) 3500 thermomechanical simulator. Tests were performed following the thermomechanical profile sketched in Figure 1, designed for producing a stable microstructure close to equilibrium conditions. Specimens were first austenitized at 1100 °C for 3 minutes, and then cooled to 679 °C. They were held at this temperature for 10 minutes, during which formation

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Fig. 1-Thermomechanical process employed in the experiments.

of around 50 pct ferrite takes place, according to separate dilatation measurements. Finally, the specimens were heated to the deformation temperature and held there for 5 minutes before deformation took place. Deformation temperatures were varied in the range 763 °C to 825 °C with steps of around 12 °C. At temperatures higher than 825 °C, the steel dynamically recrystallizes at the chosen deformation conditions, while below 763 °C, intercritically deformed austenite does not recrystallize but just recovers.^[13] Thus, only a narrow temperature range is available to determine the static recrystallization kinetics of intercritically deformed austenite. In order to detect any temperature gradient along the lengthwise direction of the specimen during the axial compression tests, three thermocouples were spot welded on the surface of specimens, one in the middle as a control thermocouple and the other two at both ends of the specimen. ISO-T anvils^[14] were used to achieve a temperature gradient as small as possible. For the experiments reported here, the temperature deviation from the set point was generally less than 5 °C during isothermal holding. All specimens were deformed to a true strain of 0.5 at a strain rate of 0.1 s^{-1} . The specimens were kept at the deformation temperature and subjected to an extended isothermal hold in order to monitor the stress relaxation response. The Gleeble control system was adjusted in such a way that the hydraulic valve was kept absolutely still during the long relaxation tests and clamp movement either due to thermal or hydraulic effects was avoided. At the end of the relaxation experiment, the specimens were quenched using water jets. Some specimens were quenched immediately after the deformation. Ferritic-martensitic microstructures were observed in the quenched specimens using optical microscopy and etching in 5 pct nital or saturated picric acid. To determine the ferrite fraction, digital micrographs were taken and postprocessed using Leica (Cambridge, England) Qwin quantitative metallographic software.

III. RESULTS

A. Stress-Strain Curves during the Intercritical Deformation

Since stress relaxation test results depend on the amount of stored energy present in the material, it is important to ascertain that dynamic recrystallization does not occur during deformation. Stress-strain curves at intercritical temperatures were examined and are shown in Figure 2. From the figure, it is clear that at all temperatures the stress increased con-



Fig. 2—Stress-strain curves during deformation for the temperature range 763 $^{\circ}\mathrm{C}$ to 825 $^{\circ}\mathrm{C}.$

tinuously with deformation, suggesting that dynamic recrystallization did not occur during these intercritical deformations.

B. Relaxation Curves

The stress decreases during relaxation studied are shown in Figure 3 for the range of temperatures investigated. As reported previously,^[11,12,13] the curves exhibit three distinct regions. The stress decreases immediately after ending the deformation is linear with $\log(t)$, and can be described by the equation $\sigma = \sigma_0 - \alpha_0 \log (t)$. It is widely accepted that such linear softening is due to recovery,^[11,12] *i.e.*, in this case, austenitic plus ferritic recovery. For all tests, there is a subsequent acceleration in the stress decrease, which has been attributed to recrystallization of austenite.^[11,12] A previous study on this material has shown that under these specific experimental conditions, the static softening is indeed due to ferritic recovery and austenitic recrystallization.^[13] Therefore, the rapid stress decrease is due to recrystallization of the intercritically deformed austenite. The onset of the recrystallization is retarded, and the rate of stress decay decreases as the holding temperature decreases. Finally, toward the end of the relaxation period, there is a second inflection point and the stresslog (time) curves flatten out and become once again linear, according to $\sigma = \sigma_1 - \alpha_1 \log(t)$, suggesting that recovery is again dominating the softening process. The second inflection is often taken to indicate the completion of the recrystallization process.

If the fraction recrystallized is f_{Rex} , the instantaneous stress level during relaxation can be deduced by the rule of mixtures:

$$\sigma = (1 - f_{\text{Rex}}) \cdot (\sigma_0 - \alpha_0 \log(t)) + f_{\text{Rex}} \cdot (\sigma_1 - \alpha_1 \log(t))$$
[1]

Inversely, the fraction recrystallized, f_{Rex} , can be derived from the measured stress level after compensation for the contributions due to both linear softening processes according to

$$f_{\text{Rex}} = \frac{\sigma - (\sigma_0 - \alpha_0 \log (t))}{(\sigma_1 - \alpha_1 \log (t)) - (\sigma_0 - \alpha_0 \log (t))}$$
[2]

The fractions derived are then fitted to a JMAK equation, and the results and fits are shown in Figure 4. For the present



Fig. 3—Relaxation curves of stress vs log (time) at the various intercritical temperatures. Pairs of arrows indicate the inflexion points for start and finish of recrystallization, which is then employed for deriving recrystallization kinetics.



Fig. 4—Recrystallized austenite fraction at various intercritical temperatures. Data points derived from relaxation curves in Fig. 3 and lines are fittings by the JMAK equation.

Temperature	Linear Fitting for the First Part in Figure 3			Linear Fitting for the Last Part in Figure 3			Least-Squares fitting in Figure 4	
	σ_0 (MPa)	α_0 (MPa)	R	σ_1 (MPa)	α_1 (MPa)	R	$t_{0.5}$ (s)	n
825 °C	144.7	32.7	0.96	54.2	10.0	0.83	4.2	1.28
800 °C	168.7	34.8	0.99	68.3	13.9	0.94	6.2	0.91
788 °C	150.5	26.2	0.93	60.4	13.5	0.98	9.7	0.88
775 °C	141.9	31.5	0.99	69.8	16.4	0.99	12.7	0.70
763 °C	133.8	27.5	0.99	64.6	15.2	0.93	20.4	0.63

Table I. Fitting Parameters for Curves Shown in Figures 3 and 4

work on recrystallization kinetics, we use the following expression of the JMAK equation:

$$f_{\text{Rex}} = 1 - \exp\left(-0.69 \cdot \left(\frac{t}{t_{0.5}}\right)^n\right)$$
[3]

where $t_{0.5}$ is the time for accomplishment of 50 pct recrystallization. The fitting parameters *n* and $t_{0.5}$ are listed in Table I.

It should be mentioned that there is a certain level of uncertainty in the derived values of n and $t_{0.5}$, which results from two facts. One is the difficulty in determining the precise location of the inflection points between the first and second parts and that between the second and third parts. This leads to a considerable uncertainty in the calculated fraction recrystallized. The error is getting larger at lower temperatures when the inflection point between the first and second parts becomes relatively more obscure as less austenite exists. This is because the austenite recrystallizes more slowly at lower temperature. To determine the reliability levels for *n* and $t_{0.5}$, we selected three different regions for the linear fittings and determined *n* and $t_{0.5}$ for each condition. Figure 5 shows the fits for the relaxation curve at 763 °C. This procedure leads to *n* values ranging from 0.57 to 0.67 and $t_{0.5}$ values ranging from 14.2 to 26.7 seconds. The other reason for the uncertainty in the calculated kinetic parameters is the experimental scatter in the instantaneous stress level during the relaxation. The relaxation curves generally show some oscillation related to proportional, integrative, and differential (PID) settings in the various control loops, leading to oscillations in the $f_{\text{Rex}} - \log(t)$ curve, affecting the values of *n* and $t_{0.5}$. Its effect can be estimated by considering



Fig. 5—Stress relaxation curves at 763 °C with various assumed start and finish times for the austenite recrystallization, and the corresponding recrystallization curves and fitting parameters.



Fig. 6—Variations of the derived *n* and $t_{0.5}$ caused by the data scatter at 763 °C. The open data points are considered as scatter and not taken into account for fitting. (*a*) The slowest possible recrystallization curve. (*b*) The fastest possible recrystallization curve at 763 °C.

two extreme cases. One is to just take into account those data points situated in the minima of the oscillations (Figure 6(a)), which corresponds to the slowest possible recrystallization kinetics. The other is to consider only those data points located in the maximum of the oscillations, corresponding to the fastest possible recrystallization kinetics (Figure 6(b)). Such a procedure gives a variation of *n* from 0.67 to 0.72 and a corresponding variation of $t_{0.5}$ from 28.9 to 20.2 seconds. Based on the analysis of all data, the possible variation of the derived *n* is estimated at ± 0.1 for all temperatures, while the relative error of the $t_{0.5}$ values is ± 33 pct.

The values for *n* and $t_{0.5}$ with their variations are shown in Figure 7. The $t_{0.5}$ values obviously depend on temperature (Figure 7(a)). Its dependence on temperature is generally expressed by^[2–5]

$$t_{0.5} = k_0 \cdot \exp\left(Q/\mathrm{R}T\right)$$
^[4]

where k_0 is a constant and Q the activation energy for recrystallization, which can be calculated as 239 kJ/mol from the



Fig. 7—Dependence of (*a*) $t_{0.5}$ and (*b*) *n* in the JMAK equation on the intercritical temperature.

slope of log ($t_{0.5}$) vs 1/T, as shown in Figure 7(a). This value is within the variation among those that have been reported. For example, Q was calculated as 191 kJ/mol by Medina's empirical equation to take compositional influence into account;^[5] Laasraoui and Jonas^[3] and Sun and Hawbolt^[4] reported higher values of 252 and 263 kJ/mol, respectively, for the austenitic recrystallization in C-Mn steel; and Simielli *et al.*^[7] gave a lower value of 172 kJ/mol for the recrystallization of intercritical austenite in C-Mn steel.

The temperature dependence of the *n* values is shown in Figure 7(b). The *n* value at 825 °C, *i.e.*, in the fully austenitic region, is 1.28, which is quite consistent with other reported values of this exponent for the austenitic recrystallization, e.g., n = 1.0 by Laasraoui and Jonas,^[3] n = 1.6 by Karjalainen *et al.*^[1,2] and n = 2.0 by Sun and Hawbolt.^[4] However, it is surprising to find that *n* decreases significantly with intercritical

temperature, since *n* for austenitic recrystallization is generally considered to be a constant value between 1 and $2^{[1-4]}$ or to only marginally decrease with temperature.^[5]

C. Metallographic Examination of the Intercritical Microstructures

The austenite and ferrite phase fractions at the various intercritical temperatures were measured using quantitative metallography on samples quenched after 50 minutes of relaxation, and the results are listed in Table II. When the temperature increases from 763 °C to 800 °C, the ferrite fraction decreases from 26 to 9 pct. At the highest temperature of 825 °C, the ferrite fraction was indeed zero.

Optical observations of intercritical microstructures on quenched specimens confirm that austenite is indeed recrystallized during the relaxation at temperatures in excess of 763 °C (Figure 8). Just 4 seconds after deformation, fine austenite grains, resulting from recrystallization, were found near the ferrite/austenite interphase boundary, while austen-

in the Intercritical Region

Temperature	763 °C	775 °C	788 °C	800 °C	825 °C
Ferrite fraction, pct Confidence level	25.9	17.7	14.8	9.1	0
at 95 pct	±3.0	±2.7	±2.4	±2.1	0



ite grains far away from the interphase boundary were still unrecrystallized. This seems to suggest that interphase boundaries between austenite and ferrite are preferred nucleation sites for recrystallization, which has also been deduced from other observations.^[7,8] In addition, the micrographs show that the proeutectoid ferrite is not formed uniformly along the prior austenite boundaries (Figure 8). The thickness of the ferrite film layer varies between grain boundaries, and ranges from zero to tens of micrometers.

IV. DISCUSSION

As indicated in Section I, two microstructural features in the intercritical (deformed) microstructure, interphase boundaries and strain variations between individual austenite grains, might affect the recrystallization kinetics and the value of the Avrami exponent. In Sections A and B, we will present a quantitative analysis of the effects of both features to explain the temperature dependence of the Avrami exponent.

A. The Influence of Interphase Boundary Area on the Avrami Exponent

The smaller Avrami exponent at a lower intercritical temperature is possibly related to the magnitude of the interphase area, which harbors the preferred nucleation sites for recrystallization.

If it is assumed that ferrite is formed along the prior austenite boundaries with a uniform thickness, the ferrite fraction F_{α} is equal to

$$F_{\alpha} = \frac{V_{\gamma}^0 - V_{\gamma}^1}{V_{\gamma}^0} = 1 - \frac{r^3}{R^3}$$
 [5]

where R and r are the radii of the initial and remained austenite grain after a certain transformation, respectively. The austeniteferrite interface boundary area per unit volume after partial transformation, S, can be calculated for a spherical austenite grain:

$$S = \frac{4\pi r^2}{4/3\pi R^3} = \frac{3r^2}{R^3} = \frac{3R^2(1-F_\alpha)^{2/3}}{R^3} = \frac{3}{R}\left(1-F_\alpha\right)^{2/3}$$
$$(0 < F_\alpha \le 1) \quad [6]$$

Note that *S* keeps increasing with the decreasing ferrite fraction, while at $F_{\alpha} = 0$, Eq. [6] is not valid since *S* should be also equal to 0. The radius of the primary austenite grain, *R*, was measured as 25 μ m by image analysis. When the exponent and the specific boundary surface are both plotted against the ferrite fraction, it is found that both show a similar trend (Figure 9) and a possible causal connection might exist. In order to quantitatively interpret the influence of nucleation sites at the interphase boundaries on kinetics of recrystallization, the microstructural path method (MPM), which was first presented by Cahn^[15] and later extended by Vandermeer and Masumura,^[16,17] was explored.

If it is assumed that nucleation only takes place at grain boundaries, and that the rates of nucleation and growth are both constant, the following equation for recrystallization kinetics can be derived for the spherical growth of nuclei:^[15,16,17]

$$f_{\text{Rex}} = 1 - \exp\left\{-2SGt \cdot \int_{0}^{1} [1 - \exp(-Y_{e})]dx\right\} \quad [7]$$

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Fig. 9-Curves of the interphase boundary area (S) and Avrami exponent (*n*) vs ferrite fraction ($0 < F_{\alpha} \leq 1$).

where f_{Rex} is the fraction recrystallized, S is the active nucleating grain boundary per unit volume, G is the growth rate, and Y_e is^[16,17]

$$Y_e = \frac{\pi I G^2 t^3}{3} \left(1 - 3x^2 + 2x^3\right) \quad x \in [0,1]$$
 [8]

where *I* is the nucleation rate per unit area of bou I and G are both thermally activated, their temperative ence can be assumed to be as follows:

$$I = I_0 \exp\left(-\frac{Q_N}{RT}\right)$$
[9]

and

$$G = G_0 \exp\left(-\frac{Q_G}{RT}\right)$$
[10]

where I_0 and G_0 are constants, and Q_N and Q_G are the activation energy for nucleation and growth, respectively.

Since G and I are both difficult to determine, the following mathematic transformation to the relative time scale t' is applied, according to

$$t' = I^{1/3} G^{2/3} t [11]$$

Similarly, the normalized specific interphase boundary area is defined as

$$S' = \left(\frac{G}{I}\right)^{1/3} \cdot S = \left[\frac{G_0}{I_0} \exp\left(\frac{Q_N - Q_G}{RT}\right)\right]^{1/3} \cdot S \quad [12]$$

Equations [8], [11], and [12] can be inserted into Eq. [7], leading to

$$f_{\text{Rex}} = 1 - \exp\left\{-2S't' \\ \int_0^1 \left[1 - \exp\left(\frac{-\pi t'^3}{3}(1 - 3x^2 + 2x^3)\right)\right] dx\right\} [13]$$

Now, curves of log (ln $(1/(1 - f_{Rex})))$ vs log (t') can be plotted for various values for the normalized interface

[9] [9] [9] [9] boundary area
$$b$$
 (Figure 16). Note that the choice to fact the relative time t' instead of the real time t does not affect the value of the slope. The Avrami exponent is artificially determined from the average slope between the 1 and 99 pct recrystallized levels. It should be noted that the minimum Avrami exponent predicted by this analysis is 1, and the exponent increases from 1.0 to 1.56 when S' increases by a factor of 50 (Figure 10). For our experimental conditions, the relative change in interface area is much smaller (only 18 pct upon formation of the maximum amount of 26 pct ferrite formed in our experiments) than that which is necessary to induce a substantial change in the Avrami expo

respectively.

in the Avrami exponent. In addition, the absolute values of *n* derived from the experiment (0.63 < n < 1.28) do not correspond to that of the model (n > 1.0).

If site saturation instead of a constant nucleation rate is assumed, Eq. [8] changes to^[16]

$$Y_e = \pi N_s^0 G^2 T^2 (1 - x^2)$$
[14]

the 1 and 99 pct

S' increases by

where N_s^0 is the number of nuclei per unit interface boundary area. By following the same procedure as previously discussed, the influence of pre-existing nuclei number on the exponent can be simulated, which shows that the existence of more nuclei at the interface boundary leads to a decreasing slope of log $(\ln (1/(1 - f_{Rex})))$ vs log (t'), *i.e.*, to a smaller exponent. This is also consistent with other types of simulation results, for example, Monte Carlo simulation for recrystallization with heterogeneous nucleation and growth by Srolovitz et al.^[18] and Cellular Automata modeling for recrystallization with homogeneous and heterogeneous nucleation by Goetz and Seetharaman.^[19] Both of them show that more nuclei lead to a smaller Avrami exponent in the case of site saturation. In conclusion, the observed change in the Avrami exponent with decreasing temperature in the intercritical region is not likely to be due to the decreasing amount of interface area.



at the different values of S' (data points), and linear fitting over the range of 1 to 99 pct recrystallization (solid lines). The sample rates of data

points in (a), (b), and (c) of simulations are 0.1, 1, and 5 m^{-1/3} s^{2/3},

boundary area S' (Figure 10). Note that the choice to use

B. The Influence of Strain Heterogeneity on the Avrami Exponent

We will now concentrate on the strain heterogeneity in the austenite as being responsible for the decrease in n value with decreasing temperature. It has been argued that a heterogeneity in nucleation and growth behavior in a uniform material leads to a reduction in experimentally observed Avrami exponent.^[6,18,19] Thus, we may expect that a decreasing Avrami exponent with decreasing intercritical deformation temperature might be attributable to a more heterogeneous recrystallization kinetics due to the presence of ferrite. Since, at the same temperature, ferrite is softer than austenite, the strain accommodated in a certain austenite grain will depend on the local amount of ferrite surrounding the austenite grain. Furthermore, the thickness of formed ferrite layer may not be uniform even at the boundary of just one grain, which leads to the micro-strain in one part of the grain being different from that in another part. Thus, the strain distribution will become more heterogeneous with increasing ferrite fraction, as shown in Figure 11. This results in a nonhomogenous distribution of stored energy and hence the recrystallization kinetics varying spatially. There is also experimental evidence for a variation



Fig. 11—Illustration of the strain variation in austenite increasing from (*a*) to (*c*) with the increasing ferrite fraction (gray area at grain boundaries). Strains in the different austenite grains are in the following order: $\varepsilon_0 > \varepsilon_1 > \varepsilon_2 > \varepsilon_3 > \varepsilon_4 > \varepsilon_5 > \varepsilon_6 > \varepsilon_7$.

of stored energy, presented by Boas and Hargreave,^[20] who demonstrated that the microhardness of a deformed metal varied significantly, both within individual grains and from one grain to another.

We will now model the effect more quantitatively. The austenite fraction is divided into N classes according to the local amount of stored energy, with the recrystallization kinetics in each class being described by the JMAK equation

$$f_{\text{Rex}}^{i} = 1 - \exp(-k_{i}t^{n_{i}}) \quad i = 1...N$$
 [15]

Obviously, k_i and n_i depend on the local deformation conditions for that particular class. At 825 °C, in the fully austenitic region, the strain in the material is assumed to be uniform and $k_i = k = 0.1165 \text{ s}^{-n}$ and $n_i = n = 1.28$ for all *i*.

In the general case of the austenite not being deformed to the same extent in each grain, the overall recrystallized fraction in the material can be calculated by

$$f_{\text{Rex}} = \frac{\sum_{i=1}^{N} (1 - \exp(-k_i t^n))}{N}$$
[16]

Assuming k to be uniformly distributed between k_2 and k_1 , and $N = [(k_2 - k_1)/\Delta k]$, the right-hand side in Eq. [16] can be integrated as follows for N $\rightarrow \infty$:

$$f_{\text{Rex}} = \frac{\sum_{i=1}^{N} (1 - \exp(-k_i t^n)) \cdot \Delta k}{k_2 - k_1} = \frac{\int_{k_1}^{k_2} [1 - \exp(-k_i t^n)] dk}{k_2 - k_1}$$
$$= 1 + \frac{\exp(-k_2 t^n) - \exp(-k_1 t^n)}{t^n (k_2 - k_1)}$$
[17]

The value of *n* for the recrystallization kinetics of the austenite should for each class be equal to the value for the fully austenitic region, *i.e.*, n = 1.28, as *n* is known to be insensitive to temperature and deformation conditions. The maximum value of the kinetics parameter in the intercritical region, *i.e.*, k_2 , can now be assumed to be equal to *k* at 825 °C. We can now define the variation in behavior using a ratio parameter *m* defined as $m = k_2/k_1$. Inserting this in Eq. [17] leads to

$$f_{\text{Rex}} = 1 + \frac{\exp\left(-k_2 t^n\right) - \exp\left(-k_2 t^n/m\right)}{t^n (k_2 - k_2/m)}$$
[18]

Using this equation, the curves of log $(\ln (1/(1 - f_{Rex})) vs)$ log (*t*) can be plotted for different values of *m*, and then the average slopes between 1 and 99 pct recrystallized fraction can be determined as an Avrami exponent (Figure 12). Based on such calculations, the relation between *m* and the Avrami exponent *n* can be determined. This relation is shown in Figure 13. It is seen that the Avrami exponent decreases with increasing value of *m*, *i.e.*, with increasing spatial variation in recrystallization kinetics in the material. When *m* is about 600, *n* starts to decrease very slowly and becomes almost stable with increasing *m*. The minimum value of the Avrami exponent in this simulation is 0.67, which is quite close to the minimum exponent in the experiments, *i.e.*, 0.63 at 763 °C.

We will now estimate the actual variation in k for the present experimental conditions. The relation between k and $t_{0.5}$ is given by $k = 0.69 \cdot t_{0.5}^{-n}$, with n = 1.28. In the literature,



Fig. 12—The relation of log (ln $(1/(1 - f_{Rex}))$ vs log (t) are calculated at the given values of m by Eq. [18] at the sample rate of 1/s for the data point, and the average slope of these curves between 1 and 99 pct recrystallization is determined as n, the Avrami exponent.



Fig. 13—The curve of Avrami exponent n vs m, which is calculated by Eq. [18] over the range of 1 to 99 pct recrystallization.

it is shown that $t_{0.5}$ is proportional to ε^{-p} , with p varying between 3.55 and 3.81.^[3] This leads, with p = 3.55, to m being given by

$$m = \frac{k_2}{k_1} = \frac{\varepsilon_2^{n,p}}{\varepsilon_1^{n,p}} = \left(\frac{\varepsilon_2}{\varepsilon_1}\right)^{4.54}$$
[19]

Equation [19] indicates that a small variation of strain between austenite grains will lead to a large variation of recrystallization kinetics. For austenite grains that are not surrounded by ferrite, the maximum strain level in these grains is at least the bulk strain value, *i.e.*, $\varepsilon_2 \ge 0.5$. However, it is much more difficult to estimate the minimum strain value ε_1 , which is caused by the heterogeneous formation of proeutectoid ferrite. We argue that the minimum strain level is that in which the critical strain for recrystallization



Fig. 14—Avrami exponent calculated from Eqs. [18] and [20] and compared with the values derived from the stress relaxation curves.

of 1 to 5 pct^[6] is just reached. Inserting the appropriate values, these assumptions lead to the estimation $m \approx 35,000$ at 763 °C.

We will now estimate *m* as a function of the fraction of ferrite. We know that m = 1 when $F_{\alpha} = 0$, since in this model only the influence of neighboring ferrite on the recrystallization kinetics of austenite is taken into account. Furthermore, *m* can increase quite strongly with even a small increase of F_{α} . A simple mathematic function to describe *m* as a function of the ferrite fraction meeting the boundary conditions is therefore

$$m = \exp\left(C_1 \cdot F_\alpha\right)$$
 [20]

where C_1 is a constant. From the *m* value at 763 °C, C_1 can be determined as 40.4. Now, the variation of recrystallization kinetics caused by the formation of ferrite at other temperatures can be estimated by Eq. [20] and then inserted into Eq. [18]. The calculated Avrami exponent as a function of the ferrite fraction is shown in Figure 14, which also contains the experimental data. A rather good quantitative agreement between calculated and experimental values is observed, suggesting that the larger spatial variation of recrystallization kinetics is indeed the predominant reason for the decrease in Avrami exponent with increasing ferrite fraction or decreasing temperature in the intercritical region.

V. CONCLUSIONS

Stress relaxation measurements on an intercritically deformed C-Mn steel have shown that the Avrami exponent for the austenite recrystallization process decreases significantly with decreasing intercritical temperature or increasing ferrite fraction. Such behavior can be due to a reduction in interphase boundary area or to an increase in strain heterogeneity. A quantitative model based on the microstructural path method seems to indicate that the small change in interface boundary surface (18 pct) is not likely to result in the significant decrease of the Avrami exponent from 1.28 to 0.63, as observed in the experiment. On the other hand, a quantitative kinetic model based on the local strain variation between austenite grains due to differences in the amount of surrounding ferrite yielded nvalues very close to the experimental values over the full temperature regime explored. Based on the good agreement between model and experiment, local austenite strain variations are seen as the major origin for the observed decrease in Avrami exponent for decreasing temperature in the intercritical regime.

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