The influence of initial grain size and the presence of austenite particles during hot deformation on flow stress and on the static recrystallization behaviour of 3% Si steel have been reported earlier. The present paper reports the results of quantitative metallographic measurements of the effects of the processing variables of strain, strain rate and temperature of deformation, and temperature of annealing on the kinetics of static recrystallization and on the recrystallized grain size. A physically based model developed for the effects of the initial microstructural variables is applied to interpret the experimental observations.

2. Experimental Procedure

The full chemical compositions of the steels investigated and the overall experimental procedure have been reported previously. The ways in which the different initial microstructures were obtained by preliminary thermomechanical processing have also been detailed. Only the essential features relevant to the processing variables are given here.

The characteristics of the experimental materials are summarised in Table 1, in which the base material, Mat(A), was recrystallized and partially decarburized to obtain Mat(A1). Mat(A) was fully decarburized to obtain Mat(A2), and Mat(A1) was fully decarburized to obtain Mat(A3).

The carbon content of Mat(A) resulted in the presence of 16% austenite particles, located mainly at the ferrite grain boundaries, after reheating at 1100°C. The decarburized materials and Mat(B), with a lower carbon content, were single phase ferrite at all the reheating temperatures of 1100, 1000 and 900°C Most of the experiments were carried out by hot rolling, quenching and then cutting small specimens for annealing to recrystallize them. This procedure gives identical results to annealing directly after rolling and then quenching. For rolling, small slabs of thickness, \( h_0 \), given in Table 1, and width 50 mm for the A materials and 28 mm for Mat(B), were rolled in a Hille 50 experimental rolling mill with rolls of 136 mm diameter, using a peripheral roll speed of 222 mm/s. The temperature was measured by a Pyrotenax chromel-alumel thermocouple inserted in each slab with its bead at the geometrical centre of the slab. The Pyrotenax chromel-alumel thermocouple inserted in each slab with its bead at the geometrical centre of the slab. The Pyrotenax chromel-alumel thermocouple inserted in each slab with its bead at the geometrical centre of the slab.

The static recrystallization kinetics and the recrystallized grain size of 3% Si steel after hot rolling at temperatures of 900 to 1100°C to equivalent strains of 0.25 to 1.24 applied in a single pass or by two pass rolling have been investigated. When the annealing temperature was changed for a constant rolling temperature, an apparent activation energy of 230 kJ/mol, and a small increase in recrystallized grain size with increase in annealing temperature were found. Rolling at different temperatures had a surprisingly small effect. The growth rate of recrystallizing regions showed a rapid decrease with annealing time, being inversely proportional to time initially, but decreasing more rapidly at longer times. For conditions leading to long recrystallization times, growth rate fell to zero, resulting in plateaux in fraction recrystallized. Increase in equivalent strain accelerated recrystallization, but the effect after two pass rolling is complicated by a change in rolling direction between the passes. The majority of results can be interpreted by a simple model based on site saturated nucleation and negligible coarsening of recrystallized grains, but smaller strains in single pass rolling, or in the second pass of two pass rolling give large discrepancies with the model.

KEY WORDS: silicon steel; hot rolling; static recrystallization; quantitative metallography; strain; temperature; Zener–Hollomon parameter.

| Table 1. Experimental materials. |
|----------------------------------|
| Material | Si wt% | C wt% | \( d_0 \) μm | y % | \( h_0 \) mm |
| Mat (A)  | 3.16  | 0.036 | 850 16 | 30   |
| Mat (A1) | 3.16  | 0.036 | 235 16 (centre) | 14.7 |
| Mat (A2) | 3.16  | -     | 235 0 (centre) | 14.7 |
| Mat (A3) | 3.16  | -     | 850 0   | 30   |
| Mat (B)  | 3.31  | 0.016 | 233 0   | 15   |

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Table 2. Experimental rolling conditions.

| d₀, μm | y % | Tᵣ, °C | ε | Passes |
|--------|-----|--------|---|--------|
| 850    | 16  | 1100   | 0.44 | 1x10% + 1x15% |
|        | 0 and 16 | 0.52 | 2x10% |
| 16     | 0.59 | 1x30% + 1x15% |
|        | 0.67 | 1x30% + 1x20% |
| 0 and 16 | 0.82 | 2x30% |
| 16     | 0.82 | 1x51% |
| 0 and 16 | 1.09 | 1x61% |
| 0 and 16 | 1.24 | 3x30% |

| 233    | 0   | 1100   | 0.82 | 2x30% |
|        | 900 | 0.82 | 2x30% |
| 180    | 0   | 1100   | 0.25 | 1x10% |
|        | 0.52 | 1x51% |
|        | 0.82 | 1x30% |
| 1000   | 0.25 | 1x10% |
| 900    | 0.25 | 1x20% |
|        | 0.41 | 1x30% |
|        | 0.52 | 1x30% |
|        | 0.59 | 1x40% |

The rolling conditions applied to the slabs of different initial microstructure are summarised in Table 2. For the thicker slabs of coarse initial grain size, the total equivalent true strains, ε, were generally applied in 2 or 3 passes. The rolling temperature, Tᵣ, relates to the temperature at entry to the first pass, which was applied 10 s after withdrawal of the slab from the reheat furnace. In the time interval of 15 s between passes, the temperature fell as a result of air cooling. After rolling, slabs were water quenched within about 2 s. For the thinner slabs of finer grain size, most of the strains were applied in a single pass, with only limited two pass rolling. Because the rolling speed was constant, the mean strain rate of rolling only varied from about 2.5 to about 5 s⁻¹, depending on the initial slab thickness and reduction. In order to investigate the effect of strain rate of deformation on subsequent static recrystallization, plane strain compression tests were carried out on specimens of single phase material of initial grain size 180 μm at constant strain rates of 5, 0.5 and 0.05 s⁻¹, at 900°C. Specimens were water quenched and small samples were subsequently annealed at 900°C.

Specimens were quenched after annealing, sectioned on the through-thickness/longitudinal plane, mechanically polished and electrolytically etched in Morris’ reagent. Microstructures were examined optically and the fraction recrystallized was determined by point counting using a grid attached to the screen of the projection microscope and traverses close to 1/4 slab thickness and/or close to the centre-line. Recrystallized grains were nearly equiaxed, so grain sizes were measured as linear intercepts along the same traverses. In specimens without microstructural gradients before rolling, they were also measured on perpendicular traverses. Sufficient points or boundaries were counted to give less than 5% relative error in the values of volume fraction recrystallized and grain size. In a few cases, the migrating boundary area per unit volume separating recrystallized grains from unrecrystallized regions was also measured in order to obtain growth rates using the Cahn–Hagel equation. Because the unrecrystallized regions are heavily elongated, the number of boundaries per unit length must be measured in both the longitudinal and through-thickness directions. In order to observe sufficient boundaries, these measurements could only be made on the finer grained materials of uniform initial microstructures.

3. Results

The effect of rolling strain on the kinetics of recrystallization is illustrated in Fig. 1 for material of initial grain size, 180 μm, and in Fig. 2 for material of coarse initial grain size, 850 μm. It can be seen, as discussed elsewhere, that the curves in Fig. 1 are of typical sigmoidal form, whereas those in Fig. 2 show a plateau in recrystallization, except at the highest strain, when 16% austenite is present, Fig. 2(a). In all cases, increase in strain systematically accelerates recrystallization, and in the case of the coarse...
grained materials it also raises the level of the plateau. In fact, although the strain ranges in Figs. 1 and 2 are similar, most of the strains in Fig. 1 (except 0.82 at 1 100°C) were attained in a single rolling pass, whereas those in Fig. 2 involved two rolling passes (or three for a strain of 1.24). More smaller passes change the constraints in rolling and result in more spread in the slab width and lower temperatures in the later passes. A comparison was therefore made of the effect of a single pass to high strains with the two and three passes of Fig. 2, giving the results shown in Fig. 3. Clearly, single pass rolling leads to more rapid recrystallization, despite the fall in temperature between the second and third passes leading to higher flow stresses in multi-pass rolling.

The overall effects of variables on kinetics are summarized in terms of the time to 0.3 fraction of recrystallization, $t_{0.3}$, as a function of strain in Fig. 4. Results from Sakai and Ohashi\(^6\) are also shown in this figure for comparison. All the curves show a similar trend of a marked decrease in $t_{0.3}$ with increase in strain, with a decrease in slope as the level of rolling strain increases. The recrystallized grain sizes measured for the same range of initial microstructure and processing conditions are shown in Fig. 5. Again, it is clear that the grain size decreases with increasing strain, but with a decrease in slope as the strain level increases. As discussed previously,\(^2\) the size of recrystallizing grains in coarse grained materials becomes constant when the plateau of recrystallization is reached. The recrystallized grain size reported in Fig. 5 is the value at the plateau in these cases.

The kinetics of recrystallization reported in Fig. 1 are for temperatures of annealing the same as for rolling, so the deformed microstructure changes with temperature. In order to determine the effect of temperature when the deformed microstructure is identical, slabs deformed at 1 100°C were quenched and then annealed at 900, 1 000 and 1 100°C, with the resulting kinetics of recrystallization shown in Fig. 6. It is apparent that the effect of temperature is now somewhat larger, as indicated by comparison with the curves in
The temperature dependence is shown more clearly in Fig. 7(a) in which the time for 0.3 fraction of recrystallization is plotted against the reciprocal of the absolute annealing temperature. The data for 0% and 16% of recrystallized grain sizes decrease slightly as the annealing temperature decreases. If the results of rolling and annealing at the same temperature from Figs 4 and 5 are plotted in the same way against the reciprocal of the annealing temperature, as shown in Fig. 8, it can be seen by comparing the slopes with the slopes of the lines from Fig. 7 that there is a surprisingly small effect of deforming at different temperatures, with the only major deviation being to longer values of $t_{0.3}$ at 1 100°C.

As discussed previously for rolling and annealing at 900°C, the migrating boundary area per unit volume, $S_{mig}$, was determined for some of the slabs rolled and annealed at 900, 1 000 and 1 100°C. Results for a strain of 0.25, shown in Fig. 9, indicate that there is a small trend for $S_{mig}$ to increase with decrease in temperature, as expected from the decrease in recrystallized grain size, $d_{rec}$, with decrease in temperature. Fig. 8. The growth rate, $G$, derived from the values of $S_{mig}$ using the Cahn–Hagel equation, as described earlier, are shown as a function of annealing time for de-
formation and annealing at 1 100°C and 1 000°C in Fig. 10. It can be seen that most of the data fall close to a line of slope $-1$. The points in brackets are for fractions recrystalized of $\geq 0.90$, when the experimental uncertainty in $G$ increases significantly. However, the results for a strain of 0.25 all appear to follow the simple power relationship, whereas at the higher strains the values of $G$ fall systematically below the power law as annealing time increases. This trend is identical to the one observed at 900°C and discussed previously.2) The line for the power law at 1 100°C is identical to the one reported previously for 900°C, within experimental accuracy, whereas the line for 1 000°C appears to be consistently slightly higher.

In order to investigate the effects of the as-deformed structure on the recrystallization behaviour at a constant annealing temperature of 900°C, two types of experiment were carried out. In the first, the deformation was applied by plane strain compression testing at three different constant strain rates to an equivalent strain of 0.41, with the results shown in Fig. 11(a). Clearly, decreasing the strain rate, which leads to a reduction in the flow stress,1) significantly retards recrystallization. In the second type of experiment, two pass rolling was applied at 1 100°C and 900°C to an equivalent strain of 0.82, the slabs were quenched and subsequently annealed at 900°C, with the results shown in Fig. 11(b). In this case, there is a surprisingly small effect of the rolling temperature on the recrystallization kinetics. In order to compare the results of the two types of experiment, the Zener–Hollomon parameter

$$Z = \dot{\varepsilon} \exp \frac{Q_{\text{def}}}{RT}$$

was calculated from the equivalent strain rates, $\dot{\varepsilon}$, s$^{-1}$, and the rolling or plane strain deformation temperature, $T_d$, the experimentally determined value of activation energy for deformation,1) $Q_{\text{def}} = 315$ kJ/mol, and the gas constant $R = 8.31$ J/mol K. The results are shown in Fig. 12(a). It can be seen that difference in strain rate appears to have a larger effect than changing the rolling temperature, but it should be noted that the deformation conditions are also different. In plane strain compression, the strain of 0.41 was applied in a single constant strain rate deformation, whereas rolling to a strain of 0.82 was applied in two sequential passes of 30% reduction separated by a 15 s interval. The same comments apply to the measured recrystallized grain sizes shown in Fig. 12(b).

4. Discussion

In discussing the effects of the processing variables on recrystallization, it is useful to summarise the conclusions from the study of the effects of the microstructural variables of initial grain size and austenite content.23 Overall, it
was found that the microstructural variables have a major effect on the nucleation density, i.e., on recrystallized grain size, but only a second order effect on the growth rate. The growth rate, \(G\), decreased with time, \(t\), of isothermal annealing according to the relationship

\[ G = G_0 \left( 1 + C_2 \left( \frac{b}{T} c_j \frac{D_s}{M} G_0 \right)^{-1} \right) \] ...........................(2)

and the overall kinetics of recrystallization, quantified in terms of the time for 0.3 fraction of recrystallization, \(t_{0.3}\), could therefore be summarised by the relationship

\[ \log t_{0.3} = C_3 \left( \frac{b}{T} c_j \frac{D_s}{M} d_{\text{rex}} \right)^{1/m} \left( \frac{b}{T} c_j D_s P_\rho_0 \right) \] ...........................(3)

where \(C_2\), \(C_3\), and \(c_j\) are constants, \(b\) is Burgers vector (considered to depend on temperature as \(b = (2 \cdot 478 \pm 3 \cdot 51 \times 10^{-5} T) \times 10^{-10} \text{ m}\)), \(T\) is absolute temperature, \(D_s\) is the self diffusion coefficient, which with the density, \(c_j\), of jogs in dislocations, controls the rate of static recovery by climb, \(M\) is the mobility of the migrating boundaries between recrystallized and unrecrystallized regions, \(d_{\text{rex}}\) is the fully recrystallography grain size, and \(G_0\) is the initial growth rate of the recrystallizing regions.

\[ G_0 = M P_{\rho_0} \] ...........................(4)

where \(P_{\rho_0}\) is the driving pressure for recrystallography in the as-deformed structure. This depends on both the dislocation density inside subgrains and on the density of dislocations in subgrain (micro band) boundaries. The former contribution is considered to be dependent only on the flow stress, \(\sigma\), whereas the latter depends on the subgrain size, \(\delta\), which is an inverse function of \(\sigma\), and on the misorientation across the subgrain boundaries, \(\theta\), which increases with strain, \(\varepsilon\), even after the flow stress has reached a steady state value.\(^2\)

### 4.1. Growth Rate

When the second term in Eq. (2) is much greater than unity, Eq. (2) becomes

\[ G = \frac{1}{C_2} \left( \frac{b}{T} c_j \frac{D_s}{M} G_0 \right)^{-1} \] ...........................(5)

which is a power law, \(G = t^{-1}\), with the proportionality constant independent of \(G_0\) and determined essentially by the activation energies of migrating boundary mobility, \(Q_{\text{m}}\), where

\[ M = M_0 \exp \left( \frac{Q_{m}}{RT} \right) \] ..........................(6)

and of self diffusion, \(Q_s\), where

\[ D_s = D_0 \exp \left( \frac{Q_s}{RT} \right) \] ..........................(7)

Combining Eq. (7) with any temperature dependence of the density of jogs leads to the activation energy, \(Q_{s}\), for recovery. Thus Eq. (5) can be written as

\[ G = C_3 \left( \frac{T}{b} \exp \left( \frac{(Q_{m} - Q_s)}{RT} \right) \right) \frac{1}{t} \] ...........................(8)

From measurements of \(G\) at 1100°C, reported in Fig. 10, and at 900°C, reported elsewhere,\(^2\) the values of \(G\) were found to be identical at any time in the power law range. Equating \(G\) and \(t\) at 1100°C and 900°C in Eq. (8) leads to the conclusion that \((Q_{m} - Q_s) = \text{10.5 kJ/mol}\). This is reasonable if both processes are diffusion controlled, but \(Q_{m}\) depends on solute diffusion rather than self diffusion, or there is a temperature dependence of the jog density in the recovery equation. On the basis of the coincidence of growth rates at 900°C and 1100°C, it is expected that the growth rate at 1000°C should be the same, so it is concluded that the apparently higher value in Fig. 10 is within experimental uncertainty for the single set of results.

**Figure 13** shows Avrami plots of the fraction recrystallographed for the experimental conditions studied, including the three strains of 0.25, 0.52 and 0.82, for which the growth rates at 1100°C are shown in Fig. 10. It can be seen that, when \(G \ll t^{-1}\), the Avrami exponent is 1 and, when the growth rate falls below the power law, the Avrami exponent falls below 1, as expected for site saturated nucleation of recrystallographization. Increase in recrystallographization time in material of coarse initial grain size leads to plateaux levels that decrease as the recrystallographization time increases. As discussed previously,\(^2\) the plateau in fraction recrystallographed is
considered to arise when recovery reduces the driving pressure for growth to a level at which it is balanced by the energy to create new grain boundary area. For all the recrystallization curves, the Avrami exponent for small fractions recrystallized is $n = 1/1006$, and the appearance of a plateau, and the level at which it occurs, depend systematically on the kinetics of recrystallization, i.e. the time available for recovery. This is shown in Fig. 14, in which the fraction recrystallized at the plateau, $X_p$, is plotted against the time for 0.3 fraction recrystallized, $t_{0.3}$, for all the experimental conditions studied. It can be seen that at 1 100°C $X_p$ starts to decrease from 1.0 at a lower value of $t_{0.3}$ when 16%γ is present during rolling than in the single phase, γ=0%, material. This is consistent with the finer recrystallized grain size, when austenite is present during rolling, producing a larger dragging pressure. The intermediate values of %γ show intermediate values of $X_p$, consistent with the effects on recrystallized grain size, when austenite is present during rolling, producing a larger dragging pressure. The intermediate values of %γ show intermediate values of $X_p$, consistent with the effects on recrystallized grain size. It is also apparent in Fig. 14 that $X_p$ only starts to decrease from 1.0 at longer times on annealing at 900°C than at 1 100°C, because both recovery and recrystallization are slower than at 1 100°C. It is therefore concluded that the plateaux in recrystallization are a direct result of the competition between static recovery rate and recrystallization rate arising from the combination of internal variables (original grain size and austenite content) and processing variables (rolling temperature, rolling strain, strain rate, and annealing temperature).

### 4.2. Effect of Strain

The results for $t_{0.3}$ and $d_{rec}$ in Figs. 4 and 5 are all presented in terms of equivalent strains, $e$, but the data for the finer grained materials in the strain range 0.25 to 0.59 were obtained by increasing the reduction in a single rolling pass carried out at the start rolling temperatures indicated, whereas the data for the coarser grained material in the strain range 0.52 to 0.82 were obtained by two pass rolling for the start rolling temperature of 1 100°C, and the strain of 1.24 by rolling in three passes, each with a reduction of 30%. The temperature between each pass drops by $40°C$ in the 15 s between passes, so the final pass has taken place at a higher Z than in single pass rolling, and the changing value of Z in each pass represents a complex strain history. Interpretation of the results therefore requires consideration of the effects of $e$ and the effects of $Z$ together. Considering first the effects of strain in a single rolling pass, it can be seen from Eqs. (A1), (A2) and (A3) in Ref. 2) that a direct effect should arise from the increase in grain boundary edge length ($L_v$) and in grain boundary area ($S_v$) with increasing strain. The effect of this increase on recrystallized grain size is illustrated for $S_v$ by the line for $(S_v/S_{vo})^{1/3}$ at the top of Fig. 15(a). It is clear that this line has the wrong form to account for the experimental data, shown as points in Fig. 15. It is therefore apparent that the effect of strain is dominated by the probability terms in Eq. (A1) for nucle-
ation density. By analogy with observations on aluminium alloys (e.g. Ref. 7) the effects of strain are complex:

(1) subgrain structure develops progressively with strain and, even though the flow stress is nearly constant over the strain range of interest here, i.e. the mean subgrain size is not expected to change significantly, the subgrain size distribution, which is critical to nucleation density, may continue to change well into steady state.

(2) the relative importance of the different nucleation sites is expected to change with strain, with grain corners being more important at low strains, because of the local heterogeneity of strain, and intragranular nucleation sites becoming more important at high strains as in-grain lattice curvatures develop, and the density of geometrically necessary dislocations in boundaries increases.

(3) the approximation of nucleation to site saturation conditions may not be valid at low strains.9)

As no data are available for these, and possibly other effects of strain, the experimental values of \( d_{\text{ex}} \) after single pass (except for a strain of 0.82) rolling have been correlated by an empirical equation

\[
\begin{align*}
\log_{10} t_{0.3} = - & \frac{d_\text{ex}}{m} - \frac{b}{m} \log_{10} \left( \frac{Q_0 - Q_\text{a}}{RT} \right) - \frac{Q_\text{a}}{2.3RT} \\
\end{align*}
\]

where \( \varepsilon_r \) is the relaxation strain to achieve steady state flow stress, \( \sigma \), and is related to this stress level and therefore to the Zener–Hollomon parameter, \( m \) is a constant, which recognises the higher strain needed to achieve steady state grain size, \( d_{\text{ex},s} \), than steady state stress20,11 and has a value of 4 to fit the present data. The constant \( n \) influences the shape of the curves and has a value of 1.5 for the lines shown in Fig. 15(a). It can be seen that the lines fit the data closely and show a small effect of deformation and annealing temperature at small strain, but a large effect when steady state has been achieved. This trend is consistent with the observed effect of \( Z \) in Fig. 12 for deforming in plane strain compression at constant low strain rates to a strain of 0.41, which attains steady state.

The effect of two or three pass rolling was studied on coarse grained material, which gives the added complication that recrystallization does not go to completion. To examine the effect of deformation history on nucleation of recrystallization, the measured values of \( d_{\text{ex}} \) in Fig. 5 at the recrystallization plateau, \( X_p \), were converted to the values expected for complete recrystallization, \( d_{\text{ex}}^* \), as3)

\[
d_{\text{ex}}^* = d_{\text{ex}} X_p^{1/3} \quad \text{(10)}
\]

These values are plotted against total rolling strain in Fig. 15(b). The curve for 1 100°C from Fig. 15(a), adjusted in level for the effect of the different initial grain sizes of 850 and 180 \( \mu \text{m} \),21 is shown as the chain line for the single phase material. Comparison of the curves shows that, from a microstructural point of view, the equivalent strain in two-pass rolling is not the sum of the strains applied in each pass. The effective equivalent strain level in two-pass rolling has been much reduced by the combination of static recovery during the 15 s between passes and by the difference in constraints and shear strains in single and two-pass rolling. Finite element modelling22 suggests that the differences in shear strain in single pass and two pass rolling are large, and there are also differences in constraints on lateral spread in single and two pass rolling of the relatively narrow experimental slabs. These differences in strain path, rather than interpass recovery and the temperature gradients present in the second rolling pass, are considered to be the major cause of the retardation of recrystallization, but further systematic research is required to quantify the complex effects of two-pass rolling, and their dependence on the microstructural variables of initial grain size and the presence of second phase austenite particles.

4.3. Overall Kinetics

The influence of nucleation density, characterised by \( d_{\text{ex}}^* \) and initial growth rate, \( G_0 \), on the time to 0.3 fraction of recrystallization can be characterised using Eq. (3) by rewriting it in the form

\[
\log_{10} t_{0.3} = C_\text{c} d_{\text{ex}}^- \frac{b}{T} \left( \frac{Q_0 - Q_\text{a}}{RT} \right) - \log_{10} C_\gamma \frac{b}{T} P_{s0} + \frac{Q_\text{a}}{2.3RT}
\]

where \( Q_\text{a} \) is the effective activation energy for recovery, which takes into account the activation energy for self diffusion, \( Q_\gamma \), and any temperature dependence of the jog density, \( c \). From the earlier discussion, \( (Q_0 - Q_\text{a}) = 10.5 \text{kJ/mol} \), and the value of \( C_\gamma \) can then be found from the slopes of the plots of \( \log_{10} t_{0.3} \) vs. \( d_{\text{ex}}^* \) reported previously for 1 100°C,21 with the resulting values shown in Table 3. For rolling to a strain of 0.82 at 1 100°C, the dislocation density inside subgrains, estimated as described previously2 from the flow stress, enables the driving pressure, \( P_{s0} \), at 1 100°C to be estimated as 1.33×10^9 Pa. Using this value, the intercept of the \( \log_{10} t_{0.3} \) vs. \( d_{\text{ex}}^* \) plots reported previously2 of −1.45, the apparent activation energy for recrystallization of 230 kJ/mol, Fig. 7(a), and the temperature dependence of \( d_{\text{ex}}^* \), Fig. 7(b), and of the Burgers vector, \( b \), and the shear modulus, \( \mu \), enables the constants \( C_\gamma \) and \( Q_\text{a} \) to be found. As shown in Table 3, these values are independent of the austenite content. The value of activation energy for static recovery, \( Q_\gamma \), of 315 kJ/mol, is significantly higher than the activation energy for self-diffusion, but coincides with the activation energy for deformation,1 which is determined by the temperature dependence of the dynamic recovery rate.

Using these values of the constants for all the experimental conditions studied, with values of \( P_{s0} \) calculated as described previously2 and the measured values of \( d_{\text{ex}}^* \) leads to the calculated values of \( \log_{10} t_{0.3} \) shown against the measured values in Fig. 16. It can be seen that, despite the simplifying assumptions of site saturated nucleation and no grain coarsening during recrystallization, Eq. (11) correlates most of the data points reasonably well. The exceptions are for two-pass rolling when the second pass reduction is less than 30%. The calculated values of \( t_{0.3} \) for lower second pass rolling reductions and for single pass rolling to a strain of 0.25 are more than an order of magnitude

| Constant | \( C_0 \), \( \text{Km}^{-1} \) | \( C_1 \), \( \text{KmN}^{-1} \) | \( Q_\text{a} \), \( \text{kJ/mol} \) |
|----------|------------------|------------------|------------------|
| 0%γ      | \( 1.30 \times 10^7 \) | \( 1.16 \times 10^{12} \) | 315              |
| 16%γ     | \( 1.65 \times 10^7 \) | \( 1.16 \times 10^{12} \) | 315              |

Table 3. Values of constants in Eq. (11).
5. Conclusions

(1) Rapid static recovery during annealing reduces the driving pressure for growth of recrystallizing regions, leading to the mean growth rate, $G$, being proportional to the inverse of annealing time, $i.e., G \propto t^{-1}$.

(2) The absolute values of $G$ within the power law range of dependence on time of annealing are almost independent of the temperature of annealing, indicating that the difference in activation energy for recovery, $Q_r$, and for migration of recrystallizing boundaries, $Q_{m}$, is small. Analysis of the data leads to $(Q_r - Q_m) = 10.5 \text{ kJ/mol}$.

(3) The competition between static recovery and boundary migration rate during recrystallization leads to incomplete recrystallization when the initial microstructural variables of grain size or austenite content, or the process variables of strain and temperature of rolling, or temperature of annealing, retard recrystallization with respect to recovery. The plateau level of recrystallization, $X_p$, decreases systematically as recrystallization slows down, $i.e$, the time for 0.3 fraction of recrystallization, $t_{0.3}$, increases at a given annealing temperature.

(4) For site saturated nucleation, the nucleation density is related to the recrystallized grain size after complete recrystallization, $d_{\text{rex}}^*\text{so when a plateau occurs, the measured recrystallized grain size, } d_{\text{rex}}$, must be modified to obtain $d_{\text{rex}}^* = \frac{d_{\text{rex}}}{x^{-1/3}}$.

(5) The observed values of $d_{\text{rex}}$ for a given deformed microstructure increase with increase in annealing temperature. For single pass rolling, $d_{\text{rex}}$ decreases with increasing strain and becomes more sensitive to the deformation temperature (Zener–Hollomon parameter) as the strain increases and steady state deformed microstructures are achieved.

(6) The values of $d_{\text{rex}}$ observed after two or three pass rolling indicate that recovery (in 15 s) and a change in strain path result in the effective strain, in terms of microstructure, being much less than the sum of the equivalent strains in the two passes.

(7) A simple model based on site saturated nucleation and the effect of static recovery on growth rate of recrystallizing regions captures the effects of most of the initial microstructural variables and the process variables on the kinetics of recrystallization with reasonable accuracy, but indicates that changes in spacial distributions of stored energy and of recrystallized grain size, lead to significant deviations.

(8) Results for low strains imposed in a single rolling pass and for reductions of less than 30% in the final pass of two pass rolling indicate that the assumption of site saturated nucleation is probably invalid, but further systematic research is required to understand the observed effects.

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