

1. Introduction

Cold rolled and annealed steel sheets are one of the most important material products with an annual production of more than 200 Mton world-wide. Stringent technical requirements must be met both with respect to the bulk properties of the steel and to many aspects of surface condition. In the present review we shall be concerned only with the former. The formability of steel sheets is primarily dependent on their strength level, work hardening ability and plastic anisotropy. In most cases, these properties are dominated by the grain size and the crystallographic texture resulting from the final treatment in the batch or continuous annealing furnace. The ability to control these structures through the manufacturing process of the steel is evidently of great importance. The present paper discusses first the most significant factors that are recognised as affecting the microstructure evolution in sheet steel and then describes a computer model that has been developed to make quantitative predictions about the effect of chemistry and process variables for one particular grade, namely titanium-stabilised interstitial-free steels.

2. Annealing Phenomena in Cold Rolled Steel

The basic features of recovery, recrystallisation and grain growth in steels are similar to those in other metals as summarised recently in the text by Humphreys and Hatherly. The hardening introduced by cold working is eliminated and the ductility is increased, thereby creating the most important condition for good formability. The present paper discusses first the most significant factors that are recognised as affecting the microstructure evolution in sheet steel and then describes a computer model that has been developed to make quantitative predictions about the effect of chemistry and process variables for one particular grade, namely titanium-stabilised interstitial-free steels.

Evidence relating to mechanisms of recrystallisation and associated texture evolution in low carbon steels is briefly reviewed and some new observations are presented. These viewpoints are taken as the starting point for development of a computer model that can be used to predict the effects of steel chemistry and process parameters on the grain structures and textures of titanium-stabilised interstitial-free steel sheets. The model utilises physical principles as far as possible and combines these with empirical descriptions where necessary, fitting these with the aid of experimental data. Assumptions made in the model are clearly described.

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the texture as in Fig. 2(b). This latter orientation spread is known to be the most suitable for increasing the planar anisotropy (usually described by the Lankford coefficient or r-value) of the steel sheet which corresponds to optimisation of the deep-drawability. Much effort has therefore been devoted to controlling the evolution of the γ-fibre texture during commercial processing (e.g. Refs. 5)–8)).

Theories of recrystallisation texture formation have for many years been divided as to the respective importance of nucleation and growth. We shall not attempt to review the field here but will discuss some of the more recent findings that are relevant to the model that will be described later. In most cases there is seen to be a close similarity between the orientation spread of the early-formed grains (‘nuclei’) and the texture of the steel at completion of recrystallisation. This is demonstrated by EBSP results from selected grains in the IF steel presented in Fig. 2(c) showing that the γ-fibre texture is already well developed at an early stage of recrystallisation. This pattern is seen in the results in Fig. 3 of growing and shrinking texture components during the recrystallisation of the same IF steel. Theories based on growth selectivity invoke special relationships between growing and shrinking texture components that are not observed to be present in the recrystallised texture components {554}\{225\} lying close to the γ-fibre is favoured for growth into the deformed α-fibre components {112}\{110\}. Some clarity has recently been introduced into this confused subject through the work of Kestens and Houbaert. These authors showed that there was clear evidence for growth selection during the latter stages of recrystallisation in very heavily (95%) cold rolled ultra-low carbon steel but that none was detectable when the same steel had been rolled to 70% reduction. Apparently, the very high rolling reduction strengthens the deformation texture, and in particular the {112}\{110\} components, to a level where an influence of oriented growth becomes apparent. However, it was clear from this work that a major contribution from oriented nucleation was also present and that for the lower rolling reductions which are typical of current commercial practice, the final recrystallisation texture is dominated by oriented nucleation. Rather similar conclusions were deduced by Hashimoto et al. for an IF steel that had been rolled to 79% reduction.

Nucleation of recrystallised grains during heating takes place preferentially in local regions of heterogeneous substructure and is usually associated with specific orientations of the newly formed grains. A number of such cases are now recognised to be of importance in sheet steels.
Shear bands may occur during rolling, especially when the initial grain size is large and when carbon or nitrogen are present, dissolved as interstitial atoms. Recrystallised grains growing from shear bands demonstrate the Goss texture, \{110\} \langle 001 \rangle. \(13)\) Nucleation may take place by intra-granular subgrain growth within the dense structure of the deformed \(g\)-fibre grains. This mechanism was originally proposed more than thirty years ago\(2,15)\) and has been substantiated by more recent works employing the modern techniques of SEM/EBSP (e.g. Refs. 16–18). Large misorientations are found to exist within the deformed grains and the new grains generally grow with orientations close to the \(g\)-fibre.

At the earliest stages of recrystallisation, new grains are most often seen at or adjacent to the old grain boundaries. The precise mechanism of their formation is not clear but they are probably associated with dislocation pile-ups where high local misorientations and stored energy are concentrated. Such grain boundary sites in cold rolled iron have been shown to generate recrystallised grain having \(g\)-fibre orientations.\(19)\) It is known\(20,21)\) that a fine initial grain structure prior to rolling is advantageous with regard to the anisotropy of the annealed sheet and this can reasonably be attributed to the greater density of grain boundary sites in addition to a reduction in the frequency of heterogeneities such as deformation bands and shear bands. In fact, the specific surface area of prior grain boundary in a cold rolled steel depends on both the initial grain size and the rolling reduction since the boundaries become extended during the deformation. In Fig. 4 the average \(r\)-values for annealed ultra-low carbon steels\(21)\) have been plotted for four initial grain sizes and three levels of rolling reduction where these have been combined in a single parameter as the specific grain boundary area. The close conformity to a single relation suggests a strong role for grain boundary nucleation.

**Fig. 2.** Textures of IF steel (\(\phi_2=45^\circ\) sections) for 75% cold rolled steel, (a) as rolled (b) recrystallised, (c) orientations of early nucleated grains and (d) key to main orientations.

**Fig. 3.** Changes in volume fractions of deformed and recrystallised texture components during the process of recrystallisation of Ti-stabilised IF steel cold rolled to 75%.
• Grain boundaries can also act as nucleation sites by the strain induced boundary migration (SIBM) process. In this case, however, it is the low energy α-fibre orientations that become favoured. Although this mechanism is seen very commonly in the early stages of recrystallisation (e.g. Ref. 18), the orientations do not seem to make a strong contribution to the final recrystallised texture. A possible reason for this may be that the SIBM grains contain a residue of the dislocation substructure and this increased internal energy causes them to be swallowed up by growth of other grains during the later stages of grain growth.

• Coarse second phases such as inclusions and cementite particles generate deformation zones of enhanced dislocation density where recrystallisation is readily initiated. The orientations of these grains are virtually random, however, so they only contribute to the final texture by decreasing its sharpness.

A study of growth rates of recrystallised grains in IF steel cold rolled 75% and annealed at 648°C has recently been carried out by Magnusson et al. using the extended Cahn–Hagel method developed by Juul-Jensen. This approach measures the recrystallised fraction and interfacial surface area of recrystallised grains belonging to various texture components and thence derives a mean growth rate for each component at different stages of the transformation during isothermal annealing. In this IF steel, about half of the new grains belonged to the γ-fibre while the others were quite widely spread in orientation. Figure 5(a) compares the average growth rates of the γ-fibre grains and the others and shows that there is no significant difference, nor is there much difference between the various γ-fibre components shown in Fig. 5(b). A noteworthy feature of these results is the marked decrease in growth rate that takes place as recrystallisation progresses. A satisfactory explanation of this effect is still lacking although a number of reasons can be postulated. The same phenomenon has also been observed in pure iron and other metals and it is clear that such an effect must be included when modelling recrystallisation. The constant growth rates usually assumed in Avrami-type models are not realistic.

The final grain structure produced by commercial annealing of steel sheet is not only a result of primary recrystallisation but also of grain growth during which there is normally a further strengthening of the γ-fibre texture components (e.g. Refs. 6, 27). The extent of this depends on the temperature and time of the anneal and also on the presence of dispersed second phases that inhibit grain boundary migration by their ‘Zener drag’. In practice the phases that have a strong influence are those that are finely dispersed such as AlN, TiC or NbC, N). During annealing these particles may undergo coarsening by Ostwald ripening and also dissolve to some extent at high temperatures which permits a greater degree of grain growth. Most other particles such as oxide or sulphide inclusions or TiN are too coarse and accordingly too few to contribute much to the Zener pinning force.

3. A Model for Recrystallisation in Ti-IF Steels

The present model aims to describe the kinetics of recrystallisation as well as the grain size and texture of the steel throughout the annealing cycle, also taking account of grain growth. A schematic diagram of the steel processing route is shown in Fig 6 together with the aspects that are included in the model through the various stages of the process. The model predicts statistical descriptions of the structure such as fraction recrystallised, average grain size, grain size distribution and volume fraction of γ-fibre texture. It does not reproduce actual microstructures. Listings of the input and output parameters are given in Table 1. All of the input variables are routinely available in steelworks.
production with the exception of the grain size in the hot rolled band. This must either be measured or calculated using models based on the strip mill parameters or, alternatively, a default value can be entered if the grain size does not vary strongly within the range of normal practice. The model should be applicable to any type of annealing cycle. It was originally tuned using data relevant to fast continuous annealing lines but has produced credible results even when applied to batch annealing conditions.

3.1. Underlying Assumptions

- Recrystallised grains nucleate continuously throughout the process. However, the rate of nucleation (per unit volume of deformed matrix) is greatest initially and decreases thereafter. Grains having random, \( \{ hkl \} \), orientations and \( \gamma \)-fibre, \( \{ 111 \} \), grains that nucleate intra-granularly follow similar nucleation laws, as described by Eq. (1). These grains have a short incubation period (described by a finite recrystallised fraction, \( X_0 \)) before the process of nucleation commences. Grains of the \( \gamma \)-fibre are also considered to nucleate along prior grain boundaries with a frequency proportional to the grain boundary area, i.e. depending on the initial grain size and rolling reduction. This process commences without any incubation time and is described by Eq. (2). The nucleation rates of all three types of grains include the same exponential dependencies on strain, \( e \), and on temperature, \( T \).
- The growth rates of all grains follow the same law and are dependent on strain and temperature. The same activation energy is assumed for growth as for nucleation in accordance with observations that changes in temperature over wide ranges have almost no influence on the as-recrystallised grain size of simple steels. Growth rates of migrating fronts decrease as recrystallisation progresses (see above) with a proportionality of \( (1-x) \) where \( x \) is the fraction transformed. The growth law is described by Eq. (3). Furthermore, the growth rates of grains must take account of impingement and it is assumed that all grains have the same fraction of their surfaces ‘free to grow’ at each stage of the process.
- The calculation is carried out in small time intervals such that a new population of grains is generated in each interval and the size distributions for each type of grains are continuously logged. At any stage it is therefore possible to obtain average values for the recrystallised fraction, recrystallised grain size and volume fraction of \( \{ 111 \} \). The calculation is stopped when recrystallisation has progressed to a chosen degree of 96% and is considered to be complete.
- In its present form the model takes no account of either precipitates or elements in solid solution when considering the process of primary recrystallisation. For Ti-stabilised IF steels there is reason to believe that second phase particles have little influence at this stage.
- Nitrogen and sulphur in the steel are considered to combine with titanium as TiN and TiS. These particles are typically quite coarse and their influence on the microstructure development is ignored except to the extent that they consume some of the available titanium. The carbide phase TiC is treated as precipitating in ferrite to the level of equilibrium as given by the solubility product in Eq. (4). Precipitation of TiC is considered to commence immediately after coiling the hot rolled strip. Thereafter the particles are assumed to undergo Ostwald ripening during cooling of the hot band and during annealing of the cold rolled sheet. Dissolution is also modelled on the basis of Eq. (4). The average particle size is calculated using Wagner’s expression as given in Eq. (5). It may be thought that ignoring the precipitation process for TiC in the present model is an over-simplification but, in fact, the nature of the coarsening equation is such that the initial particle state rapidly becomes of little consequence.
- The Zener limiting grain size due to the TiC particles at

Fig. 6. Schematic description of the steel process route for the sheet products and the stages that are included in the present model.

Table 1. Input and output parameters of the model.

<table>
<thead>
<tr>
<th>INPUT</th>
<th>OUTPUT</th>
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<tbody>
<tr>
<td>1. Ti content</td>
<td>1. Increase in recrystallised fraction with temperature/time</td>
</tr>
<tr>
<td>2. C content</td>
<td>2. Average grain size at the end of primary recrystallisation</td>
</tr>
<tr>
<td>3. S content</td>
<td>3. Grain size distributions for random texture grains and ( { 111 } ) grains</td>
</tr>
<tr>
<td>4. N content</td>
<td>4. Volume fraction of ( { 111 } ) texture at end of recrystallisation</td>
</tr>
<tr>
<td>5. Initial grain size in hot band, ( d_{00} )</td>
<td>5. Increase in average grain size during grain growth with temperature/time</td>
</tr>
<tr>
<td>6. Coiling temperature</td>
<td>6. Change in the volume fraction of ( { 111 } ) texture during grain growth</td>
</tr>
<tr>
<td>7. Cooling rate in the coil, down to 500°C</td>
<td>7.</td>
</tr>
</tbody>
</table>
any stage is obtained from the volume fraction of this phase via Eq. (4) combined with their current average size from Eq. (5) using the expression in Eq. (7). The coefficient 0.5 in this equation was chosen in the light of experimental results and is also in good agreement with recent large scale computer simulations. The term describes a limiting grain size that would arise, for example due to other types of particle present, if the amount of TiC phase becomes very small. In most realistic situations the effect of this term is very small.

- It should be noted that the influence of TiC particles is only included in connection with grain growth after primary recrystallisation is complete. If the as-recrystallised grain size is smaller than the calculated Zener limit, then the grains are considered to grow rapidly until they reach that limit. If, on the other hand, the as-recrystallised grain size is larger than the Zener limit, then no change in average grain size occurs until coarsening of the particles raises the limit to the value of the grain size. Thereafter, the average grain size increases in step with the changing Zener limit.

- Grain growth is modelled along the lines of Hillert’s treatment. Grains larger than a critical size grow while those smaller than this size according to Hillert’s equation for normal grain growth, Eq. (8). At each stage the value of has to be evaluated by an iterative routine to ensure constancy of volume. Growth is allowed to occur until the average value of the grain size reaches the current value of the Zener limit. Both types of γ-fibre grains as well as the \( \{hkl\} \) grains are included equally in the distribution while this growth process is being calculated. However, distributions are logged separately so that the volume fraction of \([111]\) in the overall texture can be evaluated. Since the average size of \([111]\) grains is larger than the others due to the earlier nucleation of some of these at grain boundary sites, there is always a trend for the γ-fibre texture to strengthen during grain growth.

3.2. Mathematical Relationships Used in the Model

The nucleation rates (number of new grains/unit time/unit volume of deformed matrix) for random, \( \{hkl\} \), and intra-granular \([111]\) grains are described by Eq. (1). The subscript \( \gamma_m \) means γ-fibre grains nucleating within the matrix.

\[
\dot{N}_{\gamma_m} = A \cdot e^4 \cdot \exp \left( -\frac{Q}{R \cdot T} \right) \cdot \left( 1 - x \right) \quad \text{for} \quad x \geq X_0 \nonumber
\]

The constant \( A \) is different for the random and the γ-fibre grains and these values have been evaluated by fitting with experimental data. The exponent of strain equal to 4 was also determined from experiments. Similarly, the activation energy \( Q \) was found to have a value of 370 kJ/mol and neither has been adjusted subsequently. The ‘incubation period’ \( X_0 \) has also been obtained by fitting to experimental results and has a very small value, much less than 1%. A fixed value of \( X_0 \) has been used throughout.

The nucleation rate for \([111]\) grains forming adjacent to old grain boundaries, \( \dot{N}_{gb} \), is given by Eq. (2) where the constant \( B \) was found by fitting. The last term within \( \{ \} \) gives the specific surface area of prior grain boundaries in the cold rolled sheet. The hot band grain size is described by the average intercept length, \( d_{gb} \), and its expansion during straining is a close approximation to the true elliptical integral solution on the assumption that deformation is homogeneous and equal for all grains.

\[
\dot{N}_{gb} = B \cdot \varepsilon^4 \cdot \exp \left( -\frac{Q}{R \cdot T} \right) \cdot \left( 1 - x \right) \nonumber
\]

\[
\left\{ \frac{2}{d_{gb}} \cdot \left( 1 + 0.13 \cdot \varepsilon + 0.235 \cdot \varepsilon^2 + 0.200 \cdot \varepsilon^3 \right) \right\} \nonumber
\]

The growth rate during primary recrystallisation, \( G \), meaning the migration rate of boundaries separating recrystallised and deformed regions, is described by Eq. (3). This was chosen to be proportional to strain during rolling on the grounds that the dislocation density that provides the driving force for growth increases approximately linearly with strain. The activation energy is as above and the constant, \( C \), was determined by fitting with experimental data.

\[
G = C \cdot \varepsilon \cdot \exp \left( -\frac{Q}{R \cdot T} \right) \cdot \left( 1 - x \right) \nonumber
\]

The volume fraction, \( f \), of TiC precipitate and the concentration of titanium in solid solution \( T_i \), are calculated from the steel composition after deducting titanium combined as stable TiN and TiS particles, together with the solubility product of TiC as given by Eq. (4). It is assumed that the equilibrium is established at all stages following coiling and during the final anneal although as a practical simplification, no changes are calculated at temperatures below 500°C. The contents of \( T_i \) and \( C \) in Eq. (4) are in wt%.

\[
\log[T_i] \cdot [C] = -\frac{10,800}{T} + 4.45 \nonumber
\]

Coarsening of the TiC particles is assumed to follow Wagner’s equation, Eq. (5), being controlled by the rate of the slower diffusing species, titanium, in ferrite. The current particle radius at time \( t \) and the initial value \( r_0 \) is taken as zero immediately after coiling. \( \sigma \) is the specific surface energy of TiC and \( V_m \) is the molar volume for this compound. The solubility level of titanium \( T_i \) is obtained from Eq. (4). The diffusivity of titanium in ferrite, \( D \), is given by Eq. (6).

\[
r^3 - r_0^3 = \frac{8}{9} \cdot \sigma \cdot D \cdot \left( T_i \right) V_m \cdot t \nonumber
\]

\[
D = 1.42 \cdot 10^{-4} \cdot \exp \left( \frac{2.32}{R \cdot T} \right) \nonumber
\]

Based on Eqs. (4) and (5), the Zener limiting grain size is calculated as given in Eq. (7) from the volume fraction and average radius of the TiC particles. The meaning of the limiting grain size \( d_{gb} \) is the average intercept length of the ferrite grains when growth is just inhibited by the pinning particles.
During each element of grain growth the current size distribution is operated on using Hillert's equation, Eq. (8), in small increments until the average grain size reaches the Zener limit from Eq. (7). Note that this procedure is not used to determine kinetics. It allows for development of the grain size distribution and, more importantly, of the texture since the γ-fibre grains prosper at the expense of random ones. The authors are aware that this procedure is not identical to that developed in Hillert's analysis of abnormal growth but it appeared to produce a more realistic result in terms of the spread of the size distribution and the evolution of texture.

\[
\frac{1}{d_z} = 0.5 \cdot \frac{f}{r} + 1 \div 30 \tag{7}
\]

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\[
\frac{dd}{dt} = k \left( \frac{1}{d_c} - \frac{1}{d} \right) \tag{8}
\]

### 3.3. Adjustment of Free Parameters

The equations listed above include seven parameters that could not be obtained from reliable sources in the literature. These are the constants described by capital letters and the exponent of 4 in Eqs. (1) and (2). A large number of experiments were carried out on Ti-stabilised IF steels having different compositions and cold rolling reductions, subjected to a wide range of interrupted heating cycles. The scope of these experimental variables is indicated in Table 2.

Measurements of recrystallised fraction and ferrite grain size were made using scanning electron microscopy in the back-scattered channelling contrast mode (SEM-BSE). Textures were determined at completion of primary recrystallisation and after various periods of grain growth using x-ray diffraction and electron back-scattering patterns (EBSP). In most cases the latter was employed to obtain the volume fraction of γ-fibre grains, i.e. those whose {111} planes lay within 15° of the sheet plane.

Optimisation of the free parameter values was carried out by a painful iterative process, complicated by the non-linear functions and their strong interactions with one another. For example, the choice of \(X_0\) affects not only the kinetics of recrystallisation but also the grain size distribution and texture at its completion and hence the evolution of texture if grain growth occurs subsequently. The influence of one parameter has to be balanced continuously with those of the others. This optimisation process might be a fruitful application for artificial neural networks.

### 3.4. Examples of Some Applications

Figure 7 shows a comparison of the model predictions with experimental measurements of recrystallised fraction for a steel having two different rolling reductions after quenching at various temperatures with three different heating rates. In these cases recrystallisation was forced to occur during the continuous heating. However, the model has no problem in handling other situations, for example where the transformation takes place during isothermal holding or in any other stage of a thermal cycle. The kinetics of primary recrystallisation are seen to be well described.

An example of size distributions for new grains at completion of recrystallisation is given in Fig. 8. These are probably not as sharp as ones observed in reality but do show the correctly skewed shape. Notice also that there is a cut-off for the larger sizes of random and fibre texture grains at completion of primary recrystallisation.
Results of a simulation in Fig. 9 demonstrate how the process of grain growth is calculated. The as-recrystallised grain size was about 8 μm in this case while the Zener limit was slightly smaller at about 7 μm. The Zener limit is calculated to change in small increments so it appears to follow a smooth curve and when it exceeds the current value of the mean grain size by a chosen amount, the latter is then updated to equal that value of \( d_Z \). The mean grain size therefore evolves as a series of steps. The volume of the total of {111} or \( \gamma \)-fibre grains is summed and this shows how the texture also develops in an apparently stepped manner. Some rather typical results in Fig. 10 including comparisons with experimental measurements show the level of success that is achievable at present. In some cases if the content of carbon is high or if the annealing cycle is rapid or at low temperature, the Zener limit imposed by the TiC particles may never reach the value of the as-recrystallised grain size and so no grain growth can take place. Such a situation is modelled in Fig. 11.

2. Concluding Remarks

The present model may be considered as only a first stage, containing as it does many simplifying assumptions. The real complexity of the deformed microstructure is largely ignored and so no account is taken of the tendency for recrystallised grains to develop in colonies as is usually observed (e.g. Ref. 3). The description of texture by only two components, {111} and \( \{hkl\} \), is an over-simplification and the exact form of several of the equations cannot be justified rigorously. Furthermore, the model is restricted to a single type of steel product, albeit an important one.

Notwithstanding these criticisms it can be stated that the broad bases of the model are of a physical nature or are, at least, congruent with general observation. This gives some confidence that its predictions will be of the right nature even if not always rigorously correct. The main virtue of the present approach is to be able to evaluate the effect of changes in input variables and make a quantitative estimate of the result of such changes, even when several parameters are altered simultaneously. It may be used, for example, to say how to compensate in a later stage of the process such as during cold rolling or annealing for unintentional changes in steel chemistry or prior treatment. Thus the pur-
pose of the model is not to provide deeper insights into the physical processes involved, but rather to assist in the commercial production of sheet steel having desired structures and properties. Work is presently in progress to improve the approach and extend it to a wider range of steel types.

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