The $\gamma$–$\alpha$ Transformation Kinetics of Low Carbon Steels under Ultra-fast Cooling Conditions

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Current evolutions in the processing techniques of hot rolled steel sheet, like the development of an ultra fast cooling unit which cools the material after the final rolling step, stretch the domain of the material models used to control these processes. When employing a so-called ultra fast cooling unit an accurate description of the transformation kinetics is needed at cooling rates that are significantly higher than the range in which most austenite to ferrite transformation models have ever been tested. In this paper a physically based transformation model is applied to a set of dilatometer experiments involving four different commercial steel grades and cooling rates from 20 K/s to 600 K/s. The model describes the transformation by means of moving $\gamma$–$\alpha$ interfaces in a single austenite grain. Parameters that were varied in the modelling procedure are the degree of undercooling below the A3-temperature, $\Delta T$, at which the transformation is assumed to start, and the intrinsic interface mobility pre-factor, $M_0$. By analysing the errors in the fit of the calculated transformation behaviour to the experimental data, ranges of optimum fit in the $\Delta T$–$M_0$ space were determined. In order to determine a unique combination of $\Delta T$ and $M_0$, a physically justified value of $\Delta T$ was determined using the classical nucleation theory. The resulting values of $M_0$ increase with increasing cooling rate and decrease with increasing carbon content.

KEY WORDS: ultra fast cooling; phase transformation kinetics; low-carbon steel.

1. Introduction

In almost all hot strip rolling mills for low alloy steels controlled cooling of the material takes place on a so called run-out table employing laminar cooling. In such a set-up the heat is extracted by a number of laminar water jets placed at regular intervals along the length of the run-out table.1–3) The design of such cooling systems dates from the early 70s. Based on the cooling capacity of an individual water jet and on the specifications for the various controlled cooling patterns for typical steel and thickness grades, the typical length of a run-out table is of the order of 100–200 m at rolling speeds of 5–20 m/s.

Recent research has led to the development of high intensity cooling rigs in which variable cooling rates, up to 750 K/s, can be achieved.4,5) In such cooling rigs a high density of high pressure water sprays is used, with a much higher cooling power per water spray than on the conventional run-out table. Hence, the cooling section of the run-out table can be shortened considerably and cooling units currently under construction in European hot strip rolling mills only have a length of the order of 5–10 m. This intense cooling provides the possibility to produce leaner alloy steel grades with similar mechanical properties to conventional hot strip, due to the microstructural refinement accompanying the higher cooling rates.

The control of the cooling process in such a cooling unit requires great accuracy since the required temperature reduction has to be achieved in a single cooling step. This means that in order to reach the required cooling temperature within the time available, the heat released during the $\gamma$–$\alpha$ transformation has to be removed in the right quantity. If too little heat is removed, part of the transformation will occur after the cooling unit, i.e. at a lower cooling rate, which may result in relatively large ferrite grains and therefore in a loss of the otherwise improved mechanical properties. However, if too much of the heat present in the material is removed the material will leave the cooling unit at too low a temperature, resulting in the possible formation of bainite instead of ferrite. Although the yield strength of bainite is significantly higher than that of ferrite, the ductility of the material is thus reduced. Therefore this scenario is also undesirable.

In order to determine the ideal cooling profile, the kinetics of the $\gamma$–$\alpha$ transformation have to be known accurately: depending on composition and applied cooling rate the fraction ferrite as a function of temperature have to be known accurately: depending on composition and applied cooling rate the fraction ferrite as a function of temperature is needed. Predicting the transformation behaviour at such high cooling rates, immediately followed by an approximately isothermal hold simulating the effect of cooling, poses considerable additional complications due to recalescence effects, the spontaneous reheating of the material due to the release of the heat of transformation.

The present work describes a critical appraisal of a phys-
ical transformation model\(^6–^9\) originally developed for describing the transformation behaviour under conditions as encountered when cooling on a classical run-out table, with regard to conditions as encountered in an UFC-unit. The principal benefit of the present model over Avrami-type transformation equations is that it is flexible with respect to the temperature-time path. The presented transformation model determines the ferrite fraction by solving a differential equation describing the progression of the interface at every time step, while Avrami-type transformation models (e.g. Ref. 9) describe the transformation by an integral equation in which no variation in temperature-time path is allowed.

The testing of the model under these stringent requirements is performed under some simplifying assumptions for the calculations, of which the cooling-rate dependent transformation start temperature and the number of active nuclei per grain are the most important. The performance of the model has been tested for four commercial steel grades. The chosen grades are quite suitable for model testing as the ferritic microstructure remains more or less allotriomorphic even at high cooling rates, and strong recalescence related to the pearlite formation does not occur.

2. Experimental

In this study dilatometric measurements have been performed on four commercial steel grades. The chemical analysis of the alloys is summarised in Table 1. Note that the alloys are labelled A, B, C and D with increasing carbon and manganese content, and that alloy C is the sole alloy containing niobium. The temperature scheme in the measurements was: austenitisation at 1 223 K (950°C) for 10 min and cooling at a cooling rate \(\beta\) to an isothermal hold temperature \(T_{\text{hold}}\). The cooling rate \(\beta\) is varied from 20 K/s to 600 K/s, and the isothermal hold temperature \(T_{\text{hold}}\) is varied from 773 K to 973 K in steps of 50 K. The actual cooling rates of the five experiments with a nominal cooling rate of 20 K/s vary somewhat since they have been adjusted such that the cooling time between the austenitisation temperature and the hold temperature was always 15 s. The 20 K/s experiments were performed using solid samples with a diameter of 5 mm and a length of 10 mm. All other experiments were performed using hollow samples with a diameter of 3 mm, a wall thickness of 0.25 mm and a length of 10 mm. The temperature is recorded by an S-type thermocouple spot-welded to the specimen. Due to the high cooling rates a temperature gradient in the sample is to be expected. In order to estimate this gradient a second thermocouple was positioned at a distance of 2.5 mm from the central thermocouple. The maximum temperature difference observed during a single experiment varies between 25 K for slow cooling and 50 K for the highest cooling rates, but was typically of the order of 15 K during the transformation. All measurements were performed on a Bähr 805 dilatometer at the laboratory of Corus Research, Development and Technology.

The translation of a dilatometer signal into a ferrite fraction is a subject often discussed (e.g. Refs. 8, 9)). Several features have to be taken into account: the enrichment of the austenite with carbon and its effect on the lattice parameter of the phase, and the lattice parameters of ferrite, pearlite and other possible phases like bainite and martensite. However, in this study, for the sake of simplicity the ferrite fraction is determined by using the lever rule. The carbon content of the steel grades used has a maximum value of 0.463 at%, leading to a pearlite fraction of approximately 12%. Analysing the dilatometer experiments using the lever rule thus results in an error in the ferrite fraction that is smaller than the estimated overall error, which justifies the use of the lever rule.\(^{10–12}\) As shown later, the overall accuracy of the obtained ferrite fractions is estimated at 15%. To provide the model with the correct parameters both the prior austenite grain size and the effective number of ferrite nuclei per austenite grain need to be known. To achieve this aim the prior austenite and the final ferrite grain sizes were determined by means of standard metallographic techniques.

The austenite and ferrite grain sizes were determined by means of optical microscopy.

3. Results

The dilatation signal displays several features encountered in the cooling curves. These features will be explained by describing the experiments performed on alloy D, which are presented in Fig. 1. Each frame in the figure shows 5 measurements, having the same nominal cooling rate, but different isothermal hold temperatures. The isothermal hold temperatures are indicated by the dashed lines. The dilatation curves are shifted along the dilatation axis in such a way that the signals in the austenitic region coincide. This can be done without affecting the results of the analysis of the dilatation signal.

Figure 1(a) shows that in these experiments the transformation starts at approximately 1 070 K, with a gradual decrease in transformation start temperature with increasing cooling rate. According to the dilatation curves of the 20 K/s experiments the transformation is almost fully completed before reaching the isothermal hold temperatures. The observed length changes during the isothermal hold are attributed primarily to the thermal transients in the sample and the push rods holding the sample in position in the dilatometer. Although the dilatation curves for the experiments with hold temperatures below the transformation-end temperature should overlap, small instrumental deviations are observed. The scatter induced by these instrumental effects in the quantitative analysis of the dilatation data gives
rise to a systematic error between the curves on the order of 10%.

The second set of experiments concerns experiments at a cooling rate of 150 K/s and is presented in Fig. 1(b). For these higher cooling rates the transformation start temperature decreases to around 1020 K. Furthermore the dilatation curves suggest that the transformation is completed during cooling for $T_{\text{hold}} \leq 873$ K, but does evolve during the isothermal hold for higher $T_{\text{hold}}$. This means that for values of $T_{\text{hold}}$ of 973 K and 923 K part of the transformation occurs during the holding time. This part of the transformation during the isothermal holding is most noticeable for the highest $T_{\text{hold}}$.

For the experiments with cooling rates of 300 K/s and 600 K/s (Figs. 1(c) and 1(d)) the transformation-start temperatures are approximately 990 K and 980 K. For these two cooling rates the experiments with $T_{\text{hold}} = 973$ K show an almost complete transformation during the isothermal hold. It can be seen in Figs. 1(b)–1(d) that on increasing the cooling rate the thermal path of this experiment deviates from the prescribed profile. Figure 1(d), for example, shows that the temperature of the sample decreases to almost 30 K below the isothermal hold temperature before the dilatometer is able to correct for the overshoot. Since the transformation generates an additional heat source, the temperature of the sample rises to about 20 K above $T_{\text{hold}}$. For the slower cooling rates this effect is less pronounced since the time scale on which the temperature control by the dilatometer is to be executed is larger. In the experiments on steel grades A, B and C this effect can not be observed since, due to their lower carbon content, these alloys transform at higher temperatures.

The aforementioned vertical parts of the dilatometer curves at the different hold temperatures (Figs. 1(a), 1(b)) change their appearance into a curl-like shape for the cooling rates of 300 K/s and 600 K/s. This deviation should be attributed to experimental errors like temperature gradients, since a curl-like shape in a dilatation curve as can be observed in Figs. 1(c) and 1(d) cannot be explained by the physics of dilatation and transformation. These effects merely show that the absolute reliability of the measurements decreases towards the cooling rate of 600 K/s. The curls in the curves increase the error in the quantitative analysis of the transformation at this cooling rate.

In Fig. 2 the transformation curves derived from the dilatometer experiments of alloy D are presented. In the analysis of the dilatometer curves the tangents to the curves, representing the thermal expansion coefficients of austenite and ferrite, were all constructed at the same temperature range of 700 K to 850 K for ferrite and 1130 K to 1180 K for austenite. Although the limits of the ferrite fraction in these graphs should be at the fractions 0 and 1, due to the instrument factors the (apparent) ferrite fraction during the ‘curl’ at the isothermal hold can exceed the value 1. This, together with the anomalies at the hold temperature, causes overshoots and undershoots of the transformation curves above a fraction of 1 and below a fraction of 0. The magnitude of the overshoot and undershoot give a reasonable estimate of the systematic error in the determined fractions, which is about 10% to 15% in the ferrite fraction. It is possible to correct for this error by downscaling the entire curve such that the ferrite fraction ranges between 0 and 1. However, since it is not known how accurate a linear correction would be, such a correction does not necessarily decrease the error along the entire curve. For this reason, it is chosen not to employ such a correction. In order to avoid
the influence of the unrealistic under- and overshoots in describing the transformations by the model, the calculations will aim to fit the relevant part of the curves only. The limits of this relevant range are selected to be ferrite fractions of 0.1 and 0.9, which are indicated in Fig. 2.

For steels A, B and C the development of the ferrite fraction with temperature is similar to that of steel D. The 20 K/s experiments have a starting temperature of the transformation of approximately 1 140 K, 1 130 K and 1 110 K respectively. With increasing cooling rates these temperatures decrease towards 1 100 K, 1 110 K, and 1 100 K for the 600 K/s experiments. Since the carbon contents of steels A, B and C are lower than for steel D, the transformation in these steels occurs at higher temperatures. This reduces the effects of the isothermal hold on the transformation kinetics. Only in the 600 K/s experiments on steel C part of the transformation occurs during the isothermal hold of 973 K. The effect however, is much less pronounced than for steel D, since the ferrite fraction is already close to one when the temperature reaches $T_{\text{hold}}$.

Table 2 presents the average austenite and ferrite grain sizes for all four steel grades. The ferrite grain size is determined for the cooling rates of 20 K/s, 150 K/s, 300 K/s and 600 K/s. The values suggest that for each alloy the number of ferrite grains growing in one austenite grain is approximately one. Militzer et al.\(^{13}\) however, found for a commercial steel grade with similar carbon content and austenite grain size, but with a higher manganese content, that at a cooling rate of 20 K/s approximately 11 ferrite nuclei per austenite grain were formed along the grain boundary. It was observed that for a cooling rate of 80 K/s this number increased to 23. Although no absolute conclusions about the nucleus density can be drawn from the observations from Militzer et al.,\(^{13}\) they do indicate that ferrite grain coarsening has occurred during or after the transformation (see also Ref. 8).\(^{13}\)

In Fig. 3 two examples of the microstructure are given, which are formed during the 600 K/s experiment on steel D with $T_{\text{hold}}=973$ K and $T_{\text{hold}}=773$ K. The microstructure from the experiment with an isothermal hold at 973 K consists mainly of allotriomorphic ferrite, whilst the microstructure resulting from the experiment with $T_{\text{hold}}=773$ K reveals a more bainitic character. For alloy A the microstructures of the 600 K/s experiments show no significant amounts of bainite irrespective of the hold temperature. For alloy B a bainitic character was only observed for $T_{\text{hold}}=873$ K.

4. Model

The aim of this study is to describe the measured transformation curves in an accurate way, thus testing the transformation model summarised below\(^{6–8}\) at higher cooling rates than those for which the model was developed and...
lated by this programme are based on the SGTE-database and depend on both temperature and composition of the two phases. For these simulations only the elements Fe, C and Mn have been taken into account. It is also assumed that the carbon redistributes completely during the transformation, while the manganese remains at its original concentration in the austenite and ferrite.

4.2. Geometry

In order to translate the calculated interface velocity into a ferrite fraction geometrical assumptions are required. In this study the austenite grain is taken to be a tetrakaidecahedron, the size of which is derived from optical microscopy (see Table 2). Inside this 14-faced polyhedron the ferrite is assumed to grow as sphere segments from its corners. Due to ferrite grain coarsening the exact number of ferrite grains growing in one austenite grain could not be detected. In the modelling the number of nuclei is kept constant at 24, for all simulations. This implies that at all 24 corners of the tetrakaidecahedron a nucleus is situated.

4.3. Nucleation

Equation (1) describes the interface velocity from the moment $\Delta G_{\text{int}} > 0$, or when the temperature enters the two-phase region. This approach assumes that the $\gamma-\alpha$ interface is present instantaneously upon entering the two-phase region. In practice however, the creation of such an interface, or the nucleation of the ferrite phase in the austenite grain, requires a certain amount of time. Therefore, since all experiments were done during cooling, the transformation start occurs at a temperature lower than the A3-temperature. This undercooling due to nucleation, $\Delta T$, is accounted for in the simulation programme by keeping $v$ zero for $T > T_{\text{A3}} - \Delta T$.

In order to determine the undercooling a nucleation model based on the classical nucleation theory is used. The equation giving the rate of nucleation $dN/dt$ within one austenite grain as a function of temperature is the summation of the nucleation rates of four different types ($i$) of nucleation sites: homogeneous nucleation sites, grain boundary nucleation sites, grain edge nucleation sites and grain corner nucleation sites. This leads to a nucleation rate given by

$$
\frac{dN}{dt} = \sum_i N_i^\alpha f^i \exp\left(-\frac{\Delta G^i_* \lambda}{kT}\right) \exp\left(-\frac{Q_D}{kT}\right) \quad \text{......(2)}
$$

where $N_i^\alpha$ is the number of $i$-type nucleation sites per austenite grain, $f^i$ the characteristic frequency of the iron atoms ($f^i = kT h$, with $h$ Planck’s constant), $Q_D$ the activation energy for grain boundary diffusion for iron atoms ($4 \times 10^{-19}$ J), $k$ Boltzmann’s constant, $\lambda$ a scaling factor ($10^{-5}$) and $\Delta G^i_*$ the activation energy for the formation of a critical nucleus, a nucleus which is large enough to be stable at an $i$-type nucleation site. The different nucleation sites all have a specific activation energy for the formation of a critical nucleus. The general equation for this activation energy is

$$
\Delta G^i_* = \frac{4}{27} \left( \varepsilon_{\gamma \alpha} - \varepsilon_{\gamma \gamma} \right)^2 \varepsilon_{\gamma} \Delta G^i_f \quad \text{......(3)}
$$

where $\gamma_{\alpha \gamma}$ is the surface free energy of the $\gamma$--$\alpha$ interface (0.6 J/m$^2$), $\gamma_{\gamma \gamma}$ the surface energy of an austenite grain boundary (0.85 J/m$^2$), $\Delta G_i$ the free energy difference between austenite and ferrite per unit of volume, and $z_1$, $z_2$, and $z_3$ are geometrical factors of which the values are given in Table 3. The number of nucleation sites is derived from the critical nucleus size of each type of nucleation site, $r^*$. This size is given by

$$r^* = \frac{2(z_1 \gamma_{\alpha \gamma} - z_2 \gamma_{\gamma \gamma})}{3z_3 \Delta G_i}$$

(4)

The number of nucleation sites for each type of site is given by the ratio of the volume available for these nuclei and the “space” occupied by a nucleus of twice the critical radius. The factor two is arbitrarily chosen and should reflect the “space” occupied by a nucleus of twice the critical radius.

### Table 3.

| $z_1$ | $z_2$ | $z_3$ | $N_i$ |
|-------|-------|-------|-------|
| Homogeneous | 13 | 0 | 4.2 | $N_i^1$ |
| Boundary | 3.7 | 1.6 | 0.48 | $N_i^2$ |
| Edge | 1.3 | 0.72 | 0.096 | $N_i^3$ |
| Corner | 0.51 | 0.18 | 0.16 | 24 |

where $N_i^j$ is the number of nucleation sites for each type of site as a function of the austenite grain size and the critical dimension.

Finally, two remarks are made about the scaling factor $\lambda$ in Eq. (2) and about the way the nucleation model is applied in this work. In the classical nucleation theory, the parameter $\lambda$ does not occur. However, the activation energy as calculated by Eq. (3) for the formation of a critical nucleus is unrealistically high. A reason behind this failure of the model may be that it neglects the inhomogeneities which by nature exist in a solid. Due to for example fluctuations in the carbon concentration, the activation energy of creating a critical nucleus may locally be significantly lower, enabling nucleation. However, it is as yet impossible to quantify such effects. Therefore a value was taken for $\lambda$ which yields a plausible nucleation rate within a reasonable undercooling below the A3-temperature. Note that for this reason, and because of neglect of for example misfit strain energies, the nucleation behaviour calculated is only an indication of the nucleation trend with varying cooling rate and composition.

In view of the limited accuracy of the calculated nucleation rates, in the simulations the nucleation is not taken to be time-dependent, but all nuclei are introduced at the temperature at which the calculations integrating Eq. (2) over time yield $N=24$. Figure 4 presents the evolution of $N$ for alloy D at four different cooling rates with an austenite grain size of 10 $\mu$m. The temperature at which the dashed line, indicating $N=24$, intersects the calculated curves is by definition the start temperature for the transformation. The difference between these temperatures and the A3-temperature is the undercooling $\Delta T$ used in the simulations of the transformation. Although all nuclei in the model are situated at the corners of the austenite grain, in the calculation of the nucleation rate all different type of nucleation sites have been taken into account. The nucleation at grain corners governs the nucleation kinetics.

### 5. Simulations

Using the model proposed in section 4, the transformation kinetics can be calculated. In order to obtain a good agreement between the measured and the simulated transformation curves two parameters are varied: the intrinsic-mobility pre-factor $M_0$ (see Eq. (1)) and the undercooling due to nucleation, $\Delta T$. In the initial fits, this last parameter is varied freely rather than using the nucleation-theory results. The quality of the agreement between the measurements and the simulations is expressed by means of the average squared error $\Psi$, which is determined for each cooling rate over five measurements with different $T_{\text{hold}}$:

$$\Psi = \frac{1}{n} \sum_{j=1}^{n} \sum_{f=0.1}^{0.9} (f_{\exp}(T, T_{\text{hold},j}) - f_{\text{sim}}(T, T_{\text{hold},j}))^2$$

(5)

where $n$ is the number of data points between measured ferrite fractions of 0.1 and 0.9, $j$ denotes the different measurements with identical cooling rates but different $T_{\text{hold}}$, and $f_{\exp}$ and $f_{\text{sim}}$ are the experimental and simulated ferrite fraction at each time step. The simulations have been done using the temperature-time profiles of each measurement, recorded by the dilatometer.

Figure 5 shows the average squared error $\Psi$ at each cooling rate for alloy D as a function of $M_0$ and $\Delta T$. For the other three alloys the same trend is observed. All four graphs clearly show that there is no unique minimum in the error, but a range of combinations of values for $M_0$ and $\Delta T$ exists that yield a minimum $\Psi$. This implies that solely based on the fit quality no unique set of $M_0$ and $\Delta T$ can be given which describes the transformation best.
The \((M_0, \Delta T)\)-ranges of minimum \(\Psi\) can be approximated by straight lines, which are presented in Fig. 6 for all four alloys. In principle, all combinations of \(M_0\) and \(\Delta T\)-values that lie on these lines give a fit of similar quality. In order to determine the physically most realistic combination of parameters \(M_0\) and \(\Delta T\), the undercooling is calculated by means of the classical nucleation theory as outlined in section 4. In Fig. 6 the thus calculated undercoolings are presented by dots, connected by a line to guide the eye. The intersections of these \(\Delta T\)-values with the lines of optimum fit quality yield the \(M_0\)-values deemed most realistic. They are given in Table 4.

6. Discussion

In Fig. 7 the values of \(M_0\) and the transformation start temperature for the four steel grades (derived from \(\Delta T\) by
The determined values of $M_0$ and $\Delta T$ for all steel grades and cooling rates, which are the $(M_0, \Delta T)$-values that give an optimum fit and are consistent with the classical nucleation theory (Fig. 6).

|                | 20 K/s | 150 K/s | 300 K/s | 600 K/s |
|----------------|--------|---------|---------|---------|
|                | $\Delta T$ (K) | $M_0$ (moll./Js) | $\Delta T$ (K) | $M_0$ (moll./Js) | $\Delta T$ (K) | $M_0$ (moll./Js) | $\Delta T$ (K) | $M_0$ (moll./Js) |
| A              | 57     | 0.24    | 65      | 0.63    | 69      | 1.1     | 74      | 2.4     |
| B              | 52     | 0.068   | 61      | 0.30    | 65      | 0.61    | 70      | 1.2     |
| C              | 51     | 0.042   | 59      | 0.13    | 63      | 0.41    | 68      | 0.73    |
| D              | 46     | 0.038   | 55      | 0.16    | 59      | 0.45    | 64      | 0.95    |

$T_{\text{start}} = T_{N3} - \Delta T$ from Table 4 are plotted as a function of the cooling rate (Figs. 7(a) and 7(b)) and carbon concentration (Figs. 7(c) and 7(d)). Figures 7(a) and 7(c) show a decreasing start temperature with increasing cooling rate and carbon content. The start temperature follows from the $D_T$-values calculated using the nucleation model and the observed trends therefore do not give additional information on the material behaviour. For the intrinsic interface mobility factor $M_0$ an increase with increasing cooling rate is observed. Although the overall trend of $M_0$ with increasing carbon content is decreasing, this decrease is not monotonic: a minimum is observed for alloy C (0.283 at% C). The values of $M_0$ for this alloy are lower than the values for the other three alloys. This can be understood from Table 1, which shows that alloy C is the only alloy containing niobium, an alloying element which is renowned for its retarding effects on the transformation kinetics. Keeping alloy C for this reason out of consideration, it can be seen that the decrease of $M_0$ with increasing carbon content is most pronounced for the highest cooling rate.

The model used to simulate the transformation makes certain assumptions about the nucleus density as well as the solute drag effects and the carbon diffusion in the austenite phase. The effects of these assumptions on the only fit parameter, $M_0$, will be discussed below.

In all simulations the nucleus density is kept constant. It has been observed experimentally however that increasing the cooling rate significantly increases the nucleus density. Therefore part of the observed increase of the value of $M_0$ with increasing cooling rate is expected to be due to the assumed constant nucleus density. As ferrite coarsening almost certainly has taken place, the true nucleation site density cannot be determined metallographically.

The influence of carbon on the transformation kinetics is mainly determined by its interface concentration in the austenite phase. This concentration is the result of the interaction between carbon diffusion in the austenite phase and the movement of the $\gamma-\alpha$ interface. Increasing the cooling rate and thus the interface velocity enlarges the difference between the actual carbon profile and the assumed carbon profile, which is flat. The same happens when the carbon content increases. Thus, the difference between the actual and apparent interface mobility increases. Taking the diffusion of carbon into account would enhance the observed in-
crease in the value of \( M_0 \) with increasing cooling rate, but reduce the observed decrease of the value of \( M_0 \) with increasing initial carbon content.

Furthermore, it is known from literature\(^{23}\) that substitutional alloying elements, like manganese, are responsible for a drag force on the interface: the solute drag. It has been found\(^{23}\) that for higher interface velocities, in this case resulting from larger cooling rates, this drag force becomes smaller. Therefore part of the increase of the interface mobility with the cooling rate could be ascribed to the decrease in solute drag.

From the above it can be concluded that the simplified treatment of microstructural effects and the neglect of solute drag effects result in an increase of the value of \( M_0 \) with increasing cooling rates. This is however partly counteracted by the decrease of the value of \( M_0 \) with increasing cooling rate by the simplified treatment of carbon diffusion during the transformation. The importance of each effect on the eventually determined relation between the value of \( M_0 \) and the cooling rate is however difficult to establish. Therefore it is yet unclear whether the observed increase of the value of \( M_0 \) with increasing cooling rate (Fig. 7) and the observed decrease of the value of \( M_0 \) with increasing carbon content are due to the assumptions made in the model, or whether these are real physical effects.

Finally, using the \( M_0 \) and \( \Delta T \) combinations presented in Table 4, the quality of fit reached in this modelling procedure is discussed. The average errors are presented in Fig. 8. Note that the errors plotted here are \( \sqrt{\Psi} \) (Eq. (5)), and therefore represent the root mean square deviations in the ferrite fractions. Figure 8 clearly shows a decrease in the average error with an increasing content of alloying elements. It also shows that the average error per alloy for all four cooling rates is approximately the same.

In Fig. 9 the results for the measurements of alloy D that are marked with an asterisk in Fig. 8 are presented. The results of the 600 K/s experiment are remarkable. Although the absolute agreement may be limited (an average error of 0.23) the simulated transformation curve shows similar recalescence features as the measurement. The measurements have a significant uncertainty which can be as large as 15%. The quality of the fits should be judged considering these experimental uncertainties. Therefore the errors in the fits found for alloys B, C and D, in the order of 10%, are deemed reasonable, whereas the error in the fits for alloy A is relatively large.

The light microscopy study performed on the transformed samples showed that during the experiments with a cooling rate of 600 K/s in some of the cases, especially for the lower isothermal hold temperatures, a microstructure has been formed with bainitic characteristics. The general shape of the grains in this structure however, was still more or less equi-axed instead of lath-like. Since the carbon content is relatively low for bainitic structures, it is not possible to conclude that the transformation occurred completely by a bainitic mechanism. For this reason, as well as since the morphology of the formed phase does not differ too much from allotriomorphic ferrite, the 600 K/s measurements were also fitted by the proposed austenite–ferrite transformation model. The fact that the values obtained for \( M_0 \) and \( \Delta T \) for this cooling rate fit in the trends observed for the lower cooling rates supports this treatment.
7. Conclusions

It can be concluded that the interface mobility/tetrakaidecahedron model for the austenite to ferrite transformation kinetics in lean carbon manganese steels, which was developed and validated for cooling rates up to 30 K/s, can also be used successfully for cooling rates up to 600 K/s, if suitable assumptions about the nucleation behaviour are made. The model is able to reproduce all experimentally observed features, even when recalescence occurs in the 600 K/s experiments. The quality of the fit of the simulations (the r.m.s. error is of the order of 10%) increases with increasing carbon content, but seems rather insensitive to the cooling rate. Using the error minimisation scheme, no unique combination was found of the undercooling below the A₃-temperature, $\Delta T$, and the intrinsic mobility pre-exponential factor, $M_0$, describing the transformation kinetics most accurately, but sets of $\Delta T$ and $M_0$ values were found to describe the experimental data equally well. However, by using classical nucleation theory physically realistic values for the undercooling as a function of the cooling rate could be determined in an independent manner. The value of the corresponding intrinsic mobility pre-exponential factor $M_0$ thus obtained increases with increasing cooling rate and decreasing carbon content. The presence of Nb in solid solution was found to reduce the value of $M_0$.

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