The Influence of Carbon Diffusion on the Character of the $\gamma$–$\alpha$ Phase Transformation in Steel

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During the austenite–ferrite phase transformation the excess amount of carbon that can not be dissolved in the growing ferrite phase accumulates in the austenite phase ahead of the moving interface. This local carbon enrichment along the interface in the austenite decreases the driving force for transformation. Two processes therefore determine the actual interface velocity: the transformation rate of the iron lattice from fcc to bcc and the diffusion of the carbon along its gradient into the austenite grain. If the diffusion of the carbon is the rate-determining process, the transformation occurs in the diffusion-controlled mode. If the rate of transformation is determined by the lattice transformation, this is called the interface-controlled or mean-field mode. When the transformation rate is not determined by a single process, this is referred to as the mixed mode of transformation.

The mixed-mode character can be identified by the carbon concentration in the austenite at the $\alpha$–$\gamma$ interface, $x^C_{\gamma\alpha}$. If $x^C_{\gamma\alpha}$ is close to the concentration given by the local $\alpha$–$\gamma$ equilibrium, the character of the transformation is predominantly diffusion controlled. On the other hand, if diffusion is relatively fast, $x^C_{\gamma\alpha}$ is close to the average concentration in the austenite, and the transformation has an interface-controlled character. Due to the build-up of concentration gradients, the transformation character can gradually change during the transformation. In this paper, a two-dimensional study of a mixed-mode transformation is performed for Fe–C alloys of different carbon contents, and the character of the transformation is quantitatively determined.

KEY WORDS: iron–carbon alloys; ferrite formation; mixed-mode phase transformations; interface control; diffusion control.

1. Introduction

After the last stage of the hot rolling process of steel sheet, the material is cooled on a run-out cooling table. This cooling step is of the utmost importance to the production process, since it determines the final microstructure of the material and thus its mechanical properties. The reason this cooling step has such a large impact on the microstructure is that during cooling the lattice structure of the material transforms from fcc (austenite, $\gamma$) to bcc (ferrite, $\alpha$). The kinetics of this transformation is determined to a large extent by the process conditions on the run-out cooling table. In order to control the $\gamma$–$\alpha$ transformation, a profound knowledge of the kinetics of the transformation in relation to conditions like cooling rate and material composition is needed.

An important alloying element in steel is carbon. To study the influence of carbon in more detail often Fe–C alloys are used as a model system. During the transformation from austenite to ferrite in an Fe–C alloy, two diffusion processes play a role: the long range diffusion of carbon atoms in the austenite phase ahead of the moving $\gamma$–$\alpha$ interface and the short range diffusion of the alloying and matrix atoms across the $\gamma$–$\alpha$ interface. The latter process causes the movement of the $\gamma$–$\alpha$ interface into the austenite and is therefore referred to as interface migration. The origin of the carbon gradients in the austenite phase lies in the low solubility of carbon in ferrite, which is about two orders of magnitude smaller than that in austenite. Therefore, the growth of the ferrite phase causes an accumulation of carbon ahead of the moving interface. The exact magnitude of this pile-up depends on the diffusion properties of carbon in the austenite phase. An excess amount of carbon at the $\gamma$–$\alpha$ interface causes a decrease in the driving force for transformation, and therewith a retardation of the moving interface. Thus it can be concluded that the migration of the interface and the diffusion of carbon in the austenite phase are coupled processes that together determine the kinetics of the $\gamma$–$\alpha$ phase transformation.

Many austenite–ferrite phase transformation models are based on the approach that has been followed by Christian. Christian states that the velocity of the moving $\gamma$–$\alpha$ interface is the product of the driving force for transformation, the Gibbs free energy difference between the two phases, and the mobility of the interface. By combining this model with a geometry for the original austenite grain...
and the newly formed ferrite grains the development of the fraction ferrite with time and/or temperature can be calculated. This has been done on different austenite geometries in one as well as in three dimensions.\textsuperscript{3-4} In most of these models the redistribution of carbon in the austenite phase is assumed to be instantaneous. This implies that the carbon concentration inside the austenite remains homogeneous throughout the transformation. The diffusion-controlled transformation models are mainly based on the classical concept by Zener.\textsuperscript{5} In this case at all stages of the transformation the carbon concentrations in austenite and ferrite at the interface are equal to the \( \gamma-\alpha \) equilibrium concentrations. In a mixed-mode transformation this is not the case. In a mixed-mode transformation the character of the transformation can be expressed by the carbon concentration in the austenite at the interface. The character of the transformation is predominantly diffusion controlled if this concentration is close to the equilibrium concentration, and predominantly interface controlled if the concentration is close to the average carbon concentration in the austenite. Evidently, the character of the transformation is dependent on the carbon concentration and the temperature, but it can also gradually change during the transformation due to the development of the carbon concentration profile.

In this paper two-dimensional mixed-mode model calculations of the influence of carbon on the character of a \( \gamma \rightarrow \alpha \) transformation are presented. The modelling is performed for Fe–C alloys of different carbon concentrations, from 0.1 to 0.4 at\%\,. In order to avoid a too strong influence of soft impingement, the ferrite fractions in the simulations are limited to approximately 0.5. The character of the transformation is quantified by a parameter \( Y \), reflecting the carbon concentration at the interface. The dependence of \( Y \) on the overall carbon concentration and the development of \( Y \) during the transformation are studied and discussed, yielding more insight in the development and the characteristics of the \( \gamma \rightarrow \alpha \) phase transformation.

2. Theory

2.1. Different Modes of Transformation

An often used approach to study the kinetics of the \( \gamma \rightarrow \alpha \) phase transformation considers the dissipation of the driving force for the transformation.\textsuperscript{6,7} According to Liu and Hillert the driving force for the transformation is the Gibbs free energy difference \( \Delta G_1 + \Delta G_2 \) as it is shown in Fig. 1. This energy, which will be released upon the transformation, will be dissipated by both the carbon diffusion (\( \Delta G_1 \)) and the interface migration (\( \Delta G_2 \)). In order to determine these two energies, three tangents to the austenite and ferrite free energy curves are constructed: (p) the tangent to the ferrite free energy curve at \( x_C^f \), the carbon concentration in the newly formed ferrite phase—mostly the equilibrium concentration in ferrite—, (q) the tangent to the austenite free energy curve at \( x_C^\gamma \), the bulk carbon concentration in the austenite, and (r) the tangent to the austenite free energy curve at \( x_C^{\gamma\rightarrow\alpha} \), the carbon concentration in the austenite at the \( \gamma \rightarrow \alpha \) interface. Since the composition of the part of the material that actually transforms at the interface is equal to that of the ferrite phase (otherwise the ferrite concentration would vary during the transformation), the free energy differences are all evaluated at this composition (see Fig. 1). \( \Delta G_1 \) is the difference between lines (q) and (r) at \( x_C^\gamma \) and \( \Delta G_2 \) is the difference between lines (p) and (r) at this composition. From the construction in Fig. 1 the three different modes of transformation can be derived: the interface controlled mode where \( \Delta G_1 = 0 \) and \( \Delta G_2 > 0 \) (\( x_C^{\gamma\rightarrow\alpha} = x_C^\gamma \)), the diffusion controlled mode where \( \Delta G_1 > 0 \) and \( \Delta G_2 = 0 \) (\( x_C^{\gamma\rightarrow\alpha} = x_C^{\alpha} \)), and the so called mixed mode where \( \Delta G_1 > 0 \) and \( \Delta G_2 > 0 \) (\( x_C^{\gamma\rightarrow\alpha} < x_C^{\gamma\rightarrow\alpha} < x_C^{\gamma\rightarrow\alpha} \)). These three modes will be discussed below in more detail.

In the case of a transformation according to the interface controlled mode lines (r) and (q) in Fig. 1 coincide and therefore \( x_C^{\gamma\rightarrow\alpha} \) is equal to \( x_C^\gamma \). In this case the interface migration process is the rate determining process, and the diffusion of carbon in austenite is infinitely fast. Under this condition the carbon concentration in the austenite is homogeneous, and consequently the carbon concentration at the interface is equal to the average carbon concentration in the austenite. For this reason this mode is often referred to as “mean field”.\textsuperscript{2,3} Modelling the transformation using the mean field assumption only requires an equation for the interface velocity, which simplifies the problem substantially.

The other extreme situation is the diffusion controlled mode. In this case the carbon concentration in the austenite at the interface is equal to the equilibrium concentration in the austenite phase, implying that lines (r) and (p) coincide, and a carbon gradient exists in the austenite near the interface. The lattice transformation is assumed to be infinitely fast relative to the carbon diffusion. This mode is often referred to as local equilibrium. In order to find the interface velocity under local equilibrium conditions only the diffusion equation has to be solved. The solution of this equation is well feasible in one dimension, which is an often used approach.\textsuperscript{5,8,9}

The third possible mode of transformation, the mixed mode, is intermediate between the former two transformation modes. The carbon concentration at the interface in the austenite has a value between the average carbon concentration in the austenite and the equilibrium carbon concentration in this phase. Therefore the velocity of the interface is determined by both the migration of the interface and the carbon diffusion in the austenite. This mode of transformation, which is probably the most realistic description of the
transformation process, is mathematically the most complex of the three. From the literature, several treatments of the problem are known, but these treatments all have a limited applicability due to, mostly, geometrical assumptions concerning the microstructure. In order to incorporate the mixed mode in the transformation model presented in this paper, a numerical approach is applied, since an analytical solution is not feasible, due to the demand of flexibility in the distribution and number of ferrite nuclei and the complicated shape of the austenite grain.

In the present paper a two-dimensional mixed mode transformation model will be presented. Using this model the transformation of austenite to ferrite is modelled for several Fe–C alloys, using the actual diffusion characteristics of carbon in austenite, in order to monitor the character of the transformation.

2.2. Derivation of Equations

Both the migration of the \( \gamma-\alpha \) interface and the diffusion along the concentration profile have a strong effect on the carbon concentration profile in the austenite phase. Since this profile, or actually the carbon concentration at the \( \gamma-\alpha \) interface, determines the momentary interface velocity, the evolution of this profile during the transformation has to be known for a realistic simulation of the transformation kinetics.

The interface migration can be considered as a source for carbon in the austenite phase. This source can be described by the flux, \( J_{\gamma\alpha} \), of carbon into the austenite caused by the movement of the interface. This flux is related to the two carbon concentrations involved and the interface velocity by

\[
J_{\gamma\alpha} = (x_C^\gamma - x_C^\alpha)w \quad \text{..................................(1)}
\]

where \( w \) is the velocity of the moving interface. This velocity can be calculated, using the driving force \( \Delta G_2 \) (see Fig. 1) and Christian’s interface velocity model by

\[
v = M_0 \exp \left[-\frac{Q}{RT}\right] \Delta G_2 \quad \text{..................................(2)}
\]

where \( M_0 \) is the intrinsic interface mobility parameter, \( Q \) the activation energy for the interface motion, which is taken as 140 kJ/mol, \( R \) the gas constant and \( T \) the temperature. Note that the value of \( \Delta G_2 \) depends on the carbon concentration at the \( \gamma-\alpha \) interface as well as on the temperature.

Combining Eqs. (1) and (2) gives a flux of carbon at the \( \gamma-\alpha \) interface that is given by

\[
J_{\gamma\alpha} = (x_C^\gamma - x_C^\alpha)M_0 \exp \left[-\frac{Q}{RT}\right] \Delta G_2 \quad \text{..................................(3)}
\]

The second process, the diffusion of the carbon along its gradient within the austenite phase, is driven by the gradient of the chemical potential of the carbon atoms in the austenite. The resulting flux \( J_D \) is given by

\[
J_D = -M_c \mu_C \nabla \mu_C \quad \text{..................................(4)}
\]

with \( M_c \) the atomic mobility of the carbon atoms and \( \mu_C \) the chemical potential of the carbon atoms in the austenite phase. Since the diffusion takes place within a single phase, which is a dilute solution, Eq. (4) can be converted into a concentration gradient dependent equation, generally known as Fick’s first law:

\[
J_D = -D \nabla \mu_C \quad \text{..................................(5)}
\]

with \( D \) the diffusivity of carbon in the austenite phase.

The simulation of a mixed mode transformation can be performed by the application of the Eqs. (3) and (5) to a typical part of the microstructure, viz. an austenite grain in which a ferrite grain has nucleated. For the calculations a two-dimensional system is subdivided into a grid of finite area elements and time is subdivided into time steps \( \Delta t \). The description of the mixed mode problem by Eqs. (3) and (5) is however not complete, since the conditions at the boundaries are still to be defined. These boundaries are the outer boundaries of the grid and the \( \gamma-\alpha \) interface. At the outer boundaries of the grid the periodic boundary condition is applied. The periodic boundary condition assumes the simulation box to be repeated in both directions, and implies that no carbon flux across the boundaries occurs. The latter condition is defined by the concentration gradient normal to the boundary being zero. This can be effectuated by defining the diffusion coefficient across the boundaries equal to zero. Since the ferrite is assumed to have its equilibrium carbon concentration, no diffusion of carbon across the \( \gamma-\alpha \) interface is allowed, except for the flux \( J_{\gamma\alpha} \). This implies that the flux \( J_{\gamma\alpha} \) is equal to zero across the interface. This is effectively the same condition as for the outer boundaries, and is also achieved by a diffusion coefficient across the boundary equal to zero.

The fluxes of Eqs. (3) and (5) are determined individually at every area element. This implies that the austenite carbon concentration is not necessarily the same along the entire \( \gamma-\alpha \) interface. Due to, for example, the geometry of the growing ferrite phase, variations in this concentration can exist. As a consequence, the local interface velocity can also vary along the \( \gamma-\alpha \) interface. However, since grain-boundary diffusion is very rapid in comparison with bulk diffusion, in the present calculations there are no significant differences in the carbon concentration along the interface.

3. Algorithm

In order to calculate the mixed mode transformation kinetics a numerical algorithm has been developed which takes both the diffusion of the carbon and the movement of the interface into account. In this section the outline of the algorithm is explained and the implications of the initial configuration on the simulation results are studied.

3.1. Outline

In this work an explicit two-dimensional scheme on a 128 by 128 square grid with periodic boundary conditions is used to describe the problem. The ferrite is supposed to grow from the centre of the grid, starting from an initial situation in which a small number of cells in the centre is designated as ferrite, and the ferrite is surrounded by an initial carbon profile.

The carbon diffusion in the austenitic phase and the movement of the interface are two processes that are evalu-
ated on different time scales. This can be understood from the stability criterion for the numerical diffusion scheme used, an explicit scheme that will be described below. Derived from this stability criterion the time step, $\Delta t$, between two evaluations is given by

$$
\Delta t = \frac{4\delta^2}{D} \quad \text{(6)}
$$

where $D$ is the largest diffusion coefficient used in the calculations, in this case the diffusion coefficient in the mean field calculations used to test the starting configuration in the next section (2.04 $\times$ 10$^{-3}$ m$^2$/s), and $\delta$ the dimension of one area element or grid cell. With a grid of 128 by 128 cells representing a size of 50 $\mu$m, $\delta$ is equal to 0.4 $\mu$m. This results in a time step between two subsequent diffusion evaluations of 3.1 $\times$ 10$^{-5}$ s. With a typical interface velocity of 0.4 $\mu$m/s this time step implies that on average after a time lap of 3.2 $\times$ 10$^{-4} \cdot \Delta t$ a cell can be expected to transform. Therefore it has proven more efficient to evaluate the transformation progress once every 10 000 diffusion steps.

Because of the large number of time steps between two transformation steps the diffusion is not evaluated in diagonal directions, but in the orthogonal directions ($x$ and $y$) only. The evolution of the carbon concentration in the austenite is calculated explicitly by means of the rate of change in carbon concentration in both the $x$- and the $y$-direction for each time step:

$$
\frac{\partial C_x(i, j, t)}{\partial t} = \frac{\partial J_x}{\partial x} = \frac{D}{\delta^2}(x^2(i-1, j, t-\Delta t) - 2x^2(i, j, t-\Delta t) + x^2(i+1, j, t-\Delta t)) \quad \text{(7a)}
$$

and

$$
\frac{\partial C_y(i, j, t)}{\partial t} = \frac{\partial J_y}{\partial y} = \frac{D}{\delta^2}(y^2(i, j-1, t-\Delta t) - 2y^2(i, j, t-\Delta t) + y^2(i, j+1, t-\Delta t)) \quad \text{(7b)}
$$

with $(i, j)$ the co-ordinates in the grid and $t$ the time. The indices $x$ and $y$ in Eqs. (7a) and (7b) refer to the change in concentration at grid position $(i, j)$ due to the diffusion flux $J_x$ in the $x$-direction and $J_y$ in the $y$-direction, respectively. The total concentration change is equal to the sum of the changes given by these two equations, according to

$$
\frac{\partial C(i, j, t)}{\partial t} = \left( \frac{\partial C_x(i, j, t)}{\partial t} \right)_x + \left( \frac{\partial C_y(i, j, t)}{\partial t} \right)_y \quad \text{(8)}
$$

The interface diffusion is well known to be much faster than the bulk diffusion.\textsuperscript{12,13} An exact quantification of the ratio between these two diffusion rates however, is difficult to establish. In order to incorporate the effect of grain-boundary diffusion in the mixed mode simulations, the diffusion coefficient along the $\gamma-\alpha$ interface is assumed to be 100 times larger than the bulk diffusion coefficient in the austenite phase.

### Table 1.

<table>
<thead>
<tr>
<th>Parameter</th>
<th>Value</th>
</tr>
</thead>
<tbody>
<tr>
<td>$D^\gamma$</td>
<td>1.14 $\times$ 10$^{-10}$ m$^2$/s</td>
</tr>
<tr>
<td>$L$</td>
<td>50 $\times$ 10$^8$ m</td>
</tr>
<tr>
<td>$\delta$</td>
<td>0.4 $\times$ 10$^{-5}$ m</td>
</tr>
<tr>
<td>$T$</td>
<td>1050 K</td>
</tr>
<tr>
<td>$M_0$</td>
<td>0.5 mol/m</td>
</tr>
<tr>
<td>$Q$</td>
<td>140 kJ/mol</td>
</tr>
<tr>
<td>$\chi$</td>
<td>0.0657 at.%</td>
</tr>
</tbody>
</table>

The migration of the $\gamma-\alpha$ interface is evaluated locally along the interface. For each ferrite cell situated at the border of the ferrite region the progression of the interface in all directions where an austenitic neighbour cell is present is calculated and added to the progression made since the ferrite cell itself transformed. A ferrite cell is situated at the border of the ferrite region if at least one of the eight neighbouring cells is austenitic. When an austenite cell has transformed, the interface can move from the borders of that cell in at maximum eight directions: $\pm x$, $\pm y$, $\pm u$ and $\pm v$, of which $x$ and $y$ are the orthogonal directions and $u$ and $v$ the diagonal directions. For every direction where the nearest neighbour is austenitic the progression of the interface is stored. In every transformation step this progression is updated, and as soon as the progression of the interface in a certain direction exceeds the dimension of one cell (the orthogonal or diagonal dimension, depending on the direction), the nearest neighbour in that direction transforms.

When an austenite cell is situated between two ferrite cells, the progression of the interface from both ferrite cells into the austenite cell is summed.

The excess carbon is moved into the non-transforming neighbouring austenite cells. The division of the excess carbon between the neighbouring cells is proportional to the difference in the original carbon concentration in the transforming cell and the actual carbon concentration in each neighbouring cell. The parameters used in the evaluation of the transformation are summarised in Table 1.

### 3.2. The Starting Configuration

In order to assess the reliability of the established geometry of the ferrite in a simulation, the influence of the grid on the growing ferrite geometry is studied. For this purpose three mean field simulations were performed starting from three different ferrite geometries: square, diamond and octagonal. In these calculations a significantly higher diffusion coefficient was used (2.04 $\times$ 10$^{-5}$ m$^2$/s) ensuring a completely flat concentration profile in the austenite. The results of these simulations show that the ferrite geometry during all three simulations evolved via a circular shape during the early stages towards an octagonal shape at ferrite fractions near 0.5. The evolution of the ferrite fraction as a function of time is, apart from minor deviations at the start of the simulation, identical for all three simulations. For the simulations dealt with in this paper an octagonal starting ferrite geometry has been taken.
Apart from the shape of the ferritic region at the start of the transformation, the carbon profile in the austenite ahead of the interface, which is imposed in the starting configuration, may also influence the kinetics during the early stages of the simulation. In order to study the influence of the initial carbon profile, the evolution of the interface concentration in the austenite for two different starting carbon profiles has been studied. These two initial profiles have a significantly different interface concentration in the austenite. Both profiles contain a similar amount of carbon. Figure 2 shows the two profiles used and the corresponding interface concentrations of carbon in the austenite during the development of the ferrite structure. The results clearly show that the influence of this significant difference in carbon profiles on the development of the carbon interface concentration as a function of the ferrite fraction is negligible for ferrite fractions larger than 0.02.

The effect of the initial carbon concentration profile on the formation rate of the ferrite as a function of time is clearly detectable. This can be seen in Figure 3, which presents the ferrite fraction as a function of time for both initial carbon profiles. The steep carbon concentration profile, for which the thick line has been used, causes a time shift of about 0.15 s with respect to the simulation performed using the flatter profile. In this study all calculations have been performed using the same initial profile, which implies that the effect of the starting profile is identical in all calculations.

### 4. Results and Discussion

In Figure 4(a) the ferrite fraction is presented as a function of time for a mixed-mode calculation with an overall carbon content of 0.1 at% and an $M_0$-value of 0.5 mol m/J s. In the same graph, the calculated interface carbon concentration in the austenite is presented as a function of time. The carbon concentration at the interface decreases slowly from the imposed initial concentration towards a constant value of 0.6 at%. During the initial part of this decrease in concent-
tration, the increase of the ferrite fraction occurs relatively slowly with respect to later stages of the transformation. This slow start can be ascribed to two factors: the high carbon interface concentration and the relatively small size of the growing interface during this stage of the transformation.

As a reference, the interface concentration is presented in Fig. 4(b) as a function of the ferrite fraction together with the equilibrium carbon concentration \( x_{C,eq} \) and the mean field concentration \( x_{C,MF} \). The latter is calculated using the overall carbon content and the ferrite fraction. The interface concentration for \( x \approx 0.2 \) is closest to the mean field concentration and seems to run parallel to this concentration. The relative location of the interface concentration between the mean field and the equilibrium concentration can be expressed by a parameter \( Y \) that gives the character of the transformation, for which the extremes are completely diffusion controlled \((Y=1)\) and completely interface controlled \((Y=0)\). The parameter \( Y \) is defined by

\[
Y = \frac{x_{C,MF} - x_{C,eq}}{x_{C,eq} - x_{C,MF}} \quad \quad \quad (9)
\]

For the calculation presented in Fig. 4(a) the mode parameter \( Y \) is presented in Fig. 4(c), showing a considerable change in the character of the transformation during the process, from largely diffusion controlled in the initial stages to nearly interface controlled at intermediate ferrite fractions. In this case a plateau value is attained at \( Y \approx 0.24 \). At large ferrite fractions soft impingement will have a significant retarding effect on the diffusion rate, and the character of the transformation will become more diffusion-controlled (i.e. \( Y \) will increase towards 1).

A calculation performed with the same value for \( M_0 \) and an overall carbon concentration of 0.2 at% is presented in Fig. 5(a). For this higher carbon concentration the development of the interface carbon concentration is slightly different. The increase in the ferrite fraction takes place more slowly than for the calculation with an overall carbon concentration of 0.1 at%. For the 0.2 at% carbon calculation a ferrite fraction of 0.4 is reached in 4.8 s, while the 0.1 at% calculation reaches this fraction in 3.1 s. This can be understood from the carbon interface concentration for the 0.2 at% calculation, which is much higher than that for the 0.1 at% calculation, since it stabilises at a concentration of 1.4 at%.

For the calculations with the same value for \( M_0 \) and overall carbon concentrations of 0.3 at% and 0.4 at% presented in Figs. 5(b) and 5(c), the observed trend continues. The calculation with an overall carbon concentration of 0.3 at% reaches a ferrite fraction of 0.4 after 8.1 s, while a ferrite fraction of 0.4 is reached in 12.2 s for the calculation with an overall carbon concentration of 0.4 at%. The carbon concentration at the interface for these last two calculations does not show an initial decrease, but rather a slight increase towards values of 1.8 and 1.9 at%, respectively. It should be noted that the initial carbon concentration at the interface was the same for all four concentrations.

In Fig. 6 the mode parameter \( Y \) is presented for the four calculations as a function of the ferrite fraction. A clear increase of the value of \( Y \) is observed with increasing overall carbon content. This implies that by increasing the overall carbon concentration the transformation kinetics are increasingly determined by diffusion instead of the lattice transformation. The dependence of \( Y \) on the overall carbon concentration decreases significantly with increasing carbon content. For all the concentrations considered a plateau value of \( Y \) is reached for ferrite fractions between approximately 0.20 and 0.45. Figure 7 shows the observed plateau values. The calculation with an overall carbon content of 0.1 at% is the only calculation with a plateau value for \( Y \) that is closer to mean field than to local equilibrium \((Y \approx 0.24)\). From Fig. 7 therefore, it can be expected that a further increase of the overall carbon concentration leads to an even closer approach to local equilibrium, in agreement with the conclusions of several experimental studies in the literature (e.g. Ref. 14)). This approach does not proceed linearly with the overall concentration. The transition between a predominantly interface-controlled transformation and a predominant diffusion-control character occurs at an overall carbon content of approximately 0.18 at%. It should be borne in mind that the behaviour of the parameter \( Y \) that is shown in Fig. 7 is obtained with the values of \( M_0 = 0.5 \text{ molm/Js} \) and an overall carbon concentration of 0.2 at% (a), 0.3 at% (b) and 0.4 at% (c).
ble in Fig. 7 will however be general. A larger value for $M$ will shift the curve to larger $Y$-values, as will lower values of $D_g$. In fact, it is the ratio $D_g/M$ that is the characteristic governing the character of the phase transformation, in combination with the overall carbon content. In a future study, the influence of the parameters involved on the character of the phase transformation will be elaborated in more detail.

The $\gamma$–$\alpha$ transformation usually occurs during cooling rather than at a constant temperature. As a consequence the ratio $D_g/M$ will gradually decrease during the transformation, since the activation energy for carbon diffusion is smaller than that for the interface mobility. This will cause a change in the character of the transformation towards diffusion control. This effect is enhanced by the increasing driving force $\Delta G$, at increasing undercooling below the A3-temperature, leading to a larger interface velocity (cf. Eq. (2)). On the other hand, at lower temperatures the equilibrium carbon concentration in austenite will increase. This will shift the transformation towards interface control, as can be derived from Fig. 7. The latter effect can be envis-aged to be less strong, and effectively the character of the phase transformation during cooling will gradually shift towards diffusion control.

5. Conclusions

Numerical simulations on mixed-mode phase transformations, involving both the carbon diffusion and the interface mobility, yield a quantitative picture of the character of the austenite to ferrite phase transformation. On a quantitative scale between fully determined by diffusion ($Y=1$) and fully determined by interface mobility ($Y=0$), it is observed that during the initial stages the character of the transformation gradually changes, especially for the lower carbon concentrations. At ferrite fractions of approximately 0.2, the $Y$-parameter reaches a plateau value that is strongly dependent of the carbon content. For the present calculations the ratio $D_g/M$ of the diffusivity and interface mobility is such that the transition from predominantly interface-controlled to predominantly diffusion-controlled occurs at approximately 0.18 at% carbon.

REFERENCES