Mixed diffusion-controlled growth of pearlite in binary steel

BY A. S. PANDIT* AND H. K. D. H. BHADESHIA

Materials Science and Metallurgy, University of Cambridge, Pembroke Street, Cambridge CB2 3QZ, UK

A kinetic theory for the diffusion-controlled growth of pearlite is presented, which accounts simultaneously for diffusion through the austenite and via the transformation front. The simplified method abandons the need for mechanical equilibrium at the phase junctions and yet is able to explain experimental data on the growth rate of pearlite. Furthermore, unlike previous analyses, the deduced value for the activation energy for the interfacial diffusion of carbon is found to be realistic when compared with corresponding data for volume diffusion.

Keywords: pearlite; volume diffusion; interface diffusion; activation energy

1. Introduction

A colony of pearlite when viewed in three dimensions consists of an interpenetrating bicrystal of ferrite and cementite (Hillert 1962; Kral et al. 2000; Graef et al. 2006). In planar sections, the phases appear as lamellae that grow at a common front with the austenite. Cementite (θ) is rich in carbon, whereas ferrite (α) accommodates very little when it is in equilibrium with either cementite or austenite (γ). It is therefore necessary for carbon to be redistributed at the transformation front. This can happen either by diffusion in the austenite in a direction parallel to the transformation front, or by the migration of solute atoms within the α/γ and θ/γ interfaces. When the mobility of the interface is large, both of these mechanisms are said to be diffusion controlled, i.e. most of the available free energy is dissipated in driving diffusion (Bhadeshia 1985).

For a given interlamellar spacing, it has long been possible to estimate the growth rate of pearlite assuming volume diffusion control (Zener 1946; Hillert 1957), and the work has been reviewed thoroughly in Hillert (1970), Ridley (1984) and Christian (2003a, b). There have, on the other hand, been reports (Frye et al. 1953; Brown & Ridley 1966; Pearson & Verhoeven 1984) that the rates calculated in this way significantly underestimate those measured, possibly because of mass transport within the transformation interface (Sundquist 1968; Hillert 1972). Such a mechanism has also been considered in the context of cellular precipitation where the cell boundary provides an easy diffusion path, with an activation energy for the boundary-diffusion coefficient that was less than half that for volume diffusion (Turnbull 1955).

*Author for correspondence (ap524@cam.ac.uk).
There have been two further explanations proposed for the faster than expected growth rate of pearlite. The first is that transformation strain enhances diffusion (Pearson & Verhoeven 1984), but this neglects the fact that pearlite forms by reconstructive transformation, in which case, transformation strains should not be significant; furthermore, it has not been necessary to invoke such an argument in the case of other reconstructive transformations where the closure between experiment and theory is satisfactory (Bhadeshia 1985).

In recent work using phase-field modelling, it has been argued that the rate of growth is enhanced by the diffusion of carbon through the ferrite, leading to a thickening of cementite behind the transformation front (Nakajima et al. 2006). The evidence for such thickening in micrographs of partially transformed specimens is not strong (e.g. fig. 4 of Darken & Fisher (1962)). It was also admitted by the authors that there are uncertainties in the phase-field method introduced by adjustable parameters and compromises necessary to ensure numerical stability. Their analysis also neglected to consider the flux through the transformation front. It is worth noting that Nakajima et al. (2006) calculated growth rates that were significantly smaller than experimental data, and hence speculated a role of transformation strain in enhancing diffusion (Pearson & Verhoeven 1984). We will show later that this is unnecessary when the role of the inevitable flux through the transformation front is taken into account.

This highlights a general problem, that most comparisons between experiment and theory have been based on assumptions of either volume or interface diffusion control; in other words, mechanisms in isolation. Fluxes through both of these processes must, in practice, contribute to diffusion, and the relative contributions from each of these mechanisms will vary with circumstances. An elegant and first attempt at dealing simultaneously with interface and volume diffusion in Fe–C alloys is due to Hashiguchi & Kirkaldy (1984). For diffusion-controlled growth, they assumed parallel mass transfer in the volume ahead of the interface and through the interface, allowing for the Gibbs–Thompson effect at both the γ/θ and γ/α boundaries, and for mechanical equilibrium at the various interfacial junctions. The result was a rather complex theory that could not be implemented without making the important approximations

— in spite of the requirement of mechanical equilibrium, the interfaces with austenite were approximated as being flat except in the close proximity of the three-phase junctions;
— the distribution coefficient describing the ratio of the composition in the austenite in contact with ferrite or cementite, and in the transformation front, was assumed to be constant, even though the interfacial energies $\sigma_{\gamma\alpha}$ and $\sigma_{\gamma\theta}$ are not expected to be identical; and
— a simplification was made that $\sigma_{\gamma\alpha} \propto \sigma_{\alpha\theta}$ and $\sigma_{\gamma\theta} \propto \sigma_{\alpha\theta}$.

While these approximations are entirely understandable, some are clearly inconsistent with the detailed theory and the whole problem might be simplified by abandoning the need for mechanical equilibrium. Indeed, it is not strictly necessary during growth when the rate of free-energy dissipation is large, for equilibrium configurations to be respected, as long as the process leads to a net reduction in free energy. This can be seen during two-dimensional grain-growth simulations assuming orientation-independent boundary energies, where
the triple junctions do not maintain 120° angles during the process of growth, as might be required by mechanical equilibrium (Suwa & Saito 1997). Another analogy is phase transformation where the chemical potential of a particular solute can increase with the passage of the interface as long as the overall free energy is reduced.

When the approximated model (Hashiguchi & Kirkaldy 1984) was fitted to experimental data, rather large $\sigma^{\alpha\theta}$ interfacial energies were obtained, and the activation energy for the boundary diffusion of carbon was deduced to be in the range of 159 920–169 925 J mol$^{-1}$, which surprisingly is greater than for volume diffusion in both ferrite and austenite. It is noteworthy that Sundquist (1968) reported an even larger activation energy for the boundary diffusion of carbon, commenting that the expected value should be much smaller; he attributed the discrepancy to a possible role of substitutional solute impurities.

The goal in the present work was to derive a simplified theory that still deals with diffusion simultaneously through both boundary and volume and to compare the data against experiments. It will be seen that a number of difficulties, including those associated with the activation energy described above, are resolved.

2. Model formulation

(a) Assumptions

— To be consistent with diffusion-controlled growth, local equilibrium is assumed to exist at the interfaces so that the chemical potentials $\mu$ of all elements are uniform there,

$$\mu_{\text{Fe}}^\gamma = \mu_{\text{Fe}}^\alpha \quad \text{and} \quad \mu_{\text{C}}^\gamma = \mu_{\text{C}}^\alpha.$$ 

It follows that the compositions where the different phases are in contact are given by tie-lines of the equilibrium Fe–C phase diagram, which was calculated using MTDATA (NPL 2006) and the TCFE database, figure 1.

— Since the kinetic theory gives the growth rate as a function of interlamellar spacing rather than a unique velocity, it is assumed that the actual spacing adopted is that which leads to a maximum in the rate of entropy production (Hashiguchi & Kirkaldy 1984), although the maximum growth-rate criterion (Zener 1946) is also considered for the sake of completeness.

— The model is created for conditions in which fluxes from diffusion within the austenite ahead of the transformation front, and that via transport through the transformation front both contribute to growth.

(b) Diffusion coefficients

The diffusion coefficient $D_V$ of carbon in austenite is strongly dependent on concentration (Darken 1949; Wells et al. 1950). A particular model that accounts for the thermodynamics and carbon–carbon interactions while representing the known data rather well is as follows (Siller & McLellan 1970; Bhadeshia 1981):

$$D_V = \frac{kT}{\hbar} \exp \left( \frac{-\Delta G^a}{kT} \right) \left( \frac{\lambda^2}{3T_m} \right) \eta.$$ (2.1)
where $k$ and $h$ are the Boltzmann and Planck constants, respectively, and

$$\eta = a_\gamma \left[ 1 + \frac{z(1 + \chi)}{1 - (0.5z + 1)\chi + [0.25z^2 + 0.5z][1 - \phi]\chi^2} + (1 + \chi)\frac{\partial a_\gamma}{\partial \chi} \right], \quad (2.2)$$

where $\chi$ is the atom fraction of carbon, $z$ is the number of octahedral interstices around a single such interstice ($z = 12$ for austenite), $\Delta G^a$ is the activation free energy, $I_{m}$ is the activity coefficient of the activated complex, $\lambda$ is an interplanar spacing in the austenite and $a_\gamma$ is the activity of carbon in austenite. The activity and its partial differential with respect to $c$ are determined using a quasichemical thermodynamic solution model (Dunn & McLellan 1970; Bhadeshia 1980). The term $\phi$ is given by

$$\phi = 1 - \exp \left( \frac{-\omega_\gamma}{kT} \right), \quad (2.3)$$

where $\omega_\gamma$ is the nearest neighbour carbon–carbon interaction energy of 8250 J mol$^{-1}$ (Dunn & McLellan 1970). Bhadeshia (1981) found $\Delta G^a/k = 21230$ K and $\ln[I_{m}/\lambda^2] = 31.84$.

Since the diffusion coefficient of carbon in austenite is strongly concentration dependent, and given that concentration gradients drive the flux, it is necessary to calculate a weighted average coefficient (Trivedi & Pound 1967)

$$\overline{D}_V = \int^{c_e^\alpha}_{c_e^\theta} \frac{D[c^\gamma, T]}{c_e^\gamma - c_e^\gamma} \, dc^\gamma, \quad (2.4)$$

where $c_e^\alpha$ and $c_e^\theta$ are the mole fractions of carbon in austenite, which is in equilibrium with ferrite and cementite, respectively.
Figure 2. Geometry of pearlite colony. The dashed arrows indicate the volume and interface diffusion processes. The thickness of the boundary is written $\delta$.

(c) Combined fluxes during pearlite growth

A model is developed here that accounts for fluxes through both the austenite and within the transformation front, on average parallel to the front. For reasons described in §1, the model abandons the notion of interfacial tensions being balanced at three-phase junctions. As in previous work, it is assumed that diffusion within the interface can be described by a single coefficient, rather than two separate values corresponding to the $\alpha/\gamma$ and $\theta/\gamma$ interfaces. Figure 2 illustrates the geometry of the pearlite colony.

The flux $J_V$ away from the ferrite (equal to that towards the cementite), through the volume of the austenite is given by

$$J_V = -\frac{A^\alpha}{V_m} \frac{D_V}{d} \frac{dc}{dx} = \frac{D_V b S^\alpha (c_{\gamma\alpha} - c_{\gamma\theta})}{V_m S^\alpha/2},$$

(2.5)

where $V_m$ is the molar volume of austenite ($7.1 \times 10^{-6} \text{ m}^3 \text{ mol}^{-1}$) and to a good approximation assumed to be the same for all the phases involved, and $A^\alpha$ is the cross-sectional area of the interface, which for a unit depth into the diagram (figure 2) is equal to $S^\alpha$, and the diffusion distance parallel to the interface, from the ferrite to the cementite is on average $S^\alpha/2$. An equation similar to the one above can be written for the boundary-diffusion flux $J_B$ of carbon through the interface between austenite and ferrite towards the cementite (Hillert 1970),

$$J_B = -\frac{A^\alpha}{V_m} \frac{D_B}{d} \frac{dc}{dx} = \frac{12D_B \delta (c_{\gamma\alpha} - c_{\gamma\theta})}{V_m S_i}.$$

(2.6)

Interfaces are created between cementite and ferrite during the growth of pearlite, thus consuming some of the free energy $\Delta G$ of transformation. All of the available free energy is consumed in this way when the spacing between lamellae reaches a critical value $S_c = 2\sigma V_m/\Delta G$, where $\sigma$ is the $\theta/\alpha$ interfacial energy per unit area. The growth rate then becomes zero, but for $S_i > S_c$, the free-energy change is reduced by a factor $(1 - S_c/S_i)$, and the concentration difference driving diffusion becomes $(c_e^{\gamma\alpha} - c_e^{\gamma\theta})(1 - S_c/S_i)$ (Zener 1946).
The total flux arriving at the $\theta/\gamma$ interface is a combination from transport through the volume of austenite and via the boundary. It follows that, for a growth velocity $v$, the material balance at the transformation front is given by

$$\frac{v S^\alpha}{V_m} (\bar{c} - c^\alpha) = \frac{v S^\theta}{V_m} (c^\theta - \bar{c}) = \frac{v S^\alpha S^\theta}{S_i V_m} (c^\theta - c^\alpha). \quad (2.7)$$

Combining equations (2.5)–(2.7) yields

$$\frac{v S^\alpha S^\theta}{S_i} (c^\theta - c^\alpha) = 2D_V (c^{\gamma \alpha} - c^{\gamma \theta}) + \frac{12 D_B \delta (c^{\gamma \alpha} - c^{\gamma \theta})}{S_i}, \quad (2.8)$$

where $\bar{c}$ is the average concentration in the austenite. The growth velocity is now isolated as follows:

$$v = \left( \frac{c^{\gamma \alpha} - c^{\gamma \theta}}{c^\theta - c^\alpha} \right) \left( 2D_V + \frac{12 D_B \delta}{S_i} \right) \frac{S_i}{S^\alpha S^\theta} \left( 1 - \frac{S_c}{S_i} \right). \quad (2.9)$$

The problems associated with using a correct boundary-diffusion coefficient have already been emphasized. It was decided therefore to deduce this using measured data on growth rate and interlamellar spacing, due to Brown & Ridley (1966), based on the more reliable method of size distributions rather than the observation of what might be the largest colony. Given that $D_V$ is well established, the only unknown then becomes $D_B$ (figure 3), from which an activation energy for interfacial diffusion during the pearlite reaction was derived to be $Q_B \approx 97$ kJ mol$^{-1}$, with

$$D_B = 8.51 \times 10^{-5} \exp \left( -\frac{96851}{RT} \right) \text{m}^2 \text{s}^{-1}. \quad (2.10)$$
It is interesting that unlike previous work (Hashiguchi & Kirkaldy 1984; Offerman et al. 2003), where the activation energy for boundary diffusion was found to be greater than for volume diffusion in both austenite and ferrite, here, $Q_B$ is bracketed between $Q_V^a = 70 \text{kJ mol}^{-1}$ and $Q_V^f = 135 \text{kJ mol}^{-1}$ (Honeycombe & Bhadeshia 1995).

The ratio of boundary to volume diffusion flux is shown as a function of temperature in figure 4; as might be expected, boundary diffusion dominates, except at the highest of transformation temperatures.

3. Spacing criteria

There are two criteria common in fixing the interlamellar spacing of pearlite; one involves the maximization of growth rate (Zener 1946), and the other, of the entropy production rate (Kirkaldy & Sharma 1980). For volume diffusion-controlled growth, $S/S_c$ is 2 and 3, and in the case of interfacial diffusion-controlled growth, it is 1.5 and 2 for the maximum growth and entropy production criteria, respectively. We now consider how these numbers are modified for the mixed volume and interface-diffusion modes.

During isothermal transformation, the rate of entropy production is (Cahn & Hagel 1962; Kirkaldy & Sharma 1980)

$$\dot{S} = \frac{v(\Delta G_{\text{net}})}{T},$$

where $\Delta G_{\text{net}} = \Delta G - 2\sigma/S_i$ is the free energy dissipated; $\Delta G$ is the total driving force for the austenite to pearlite transformation. If it is assumed that for small undercoolings below the eutectoid temperature $T_E$, the entropy change of transformation $\Delta S$ is constant, then $\Delta G = \Delta H(1 - T/T_E)$, where $\Delta H$ is the corresponding enthalpy change, then this equation becomes

$$\dot{S} = v \frac{\Delta H \Delta T}{T_E} \left(1 - \frac{S_c}{S_i}\right).$$

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Figure 4. Relative contributions of volume and boundary-diffusion fluxes during the formation of pearlite in Fe-0.8C wt% steel.
Using equation (2.9), this becomes
\[ \dot{S} = \beta \left( \frac{2 \bar{D}_V S_i + 12 D_B \delta}{S_i^2} \right) \frac{\Delta H \Delta T}{T_E} \left( 1 - \frac{S_c}{S_i} \right)^2, \] (3.3)

where \( \beta \) contains the concentration terms. We see that \( \dot{S} \) is plotted as a function of \( S_i/S_c \) in figure 5a; the maximum in \( \dot{S} \) is obtained when \( S_i/S_c \) is between 2.01 and 2.17 depending on the temperature of transformation, but independent of the interfacial energy \( \sigma \). The corresponding calculations for the maximum growth-rate criterion are shown in figure 5b; the values of \( S_i/S_c \) vary from 1.36 to 1.53 as a function of temperature. It is not surprising that the values are similar to those expected for the case where only interface diffusion occurs, since this is the dominant process at all temperatures (figure 4).

(a) Interfacial energy

It is possible to derive from the kinetic data on pearlite growth, the interfacial energy \( \sigma \) relating cementite and ferrite (Hashiguchi & Kirkaldy 1984)
\[ \sigma = \frac{1}{2} S_c \Delta G \approx \frac{S_c \Delta T \Delta H}{2T_E}, \] (3.4)

where as explained earlier, the approximation on the right-hand side is based on the assumption that the entropy of transformation is independent of temperature (Ridley 1984; Capdevila et al. 2002; Offerman et al. 2003). We have avoided this by calculating both the enthalpy and entropy changes using MTDATA (NPL 2006); see figure 6. Values of \( S_c \) can be calculated using measured interlamellar spacings from Brown & Ridley (1966) and the entropy-production calculations illustrated in figure 5a. The interfacial energy derived in this way is illustrated as a function of temperature in figure 7. The interfacial energy values calculated here vary from those of Hashiguchi & Kirkaldy (1984), though the spacing and velocity
Figure 6. Free energy, enthalpy and entropy change as a function of temperature.

Figure 7. Comparison of calculated ferrite–cementite interfacial energy values and those reported in previous work by Hashiguchi & Kirkaldy (1984).

data are, in both cases, from the same experimental measurements (Brown & Ridley 1969). This difference may be attributed to two factors:

— the $\sigma$ calculated by Hashiguchi & Kirkaldy (1984) is based on the assumption that the entropy change during the pearlite transformation is independent of temperature, whereas we have avoided this approximation, and

— the computation of $S_i/S_c$ based on the maximum entropy production rate was in the range 2.18–2.4 (Hashiguchi & Kirkaldy 1984), whereas in the present work, it has been shown graphically that this ratio lies in the range of 2.03–2.17 for the temperatures studied. This is attributed to the different growth equations used in the two studies.

Independent, published measurements of $\sigma$ are presented in table 1. Some are estimated on the basis of studies of the coarsening of cementite in ferrite (Deb & Chaturvedi 1982; Das et al. 1993), from dihedral angle measurements...
Figure 8. Temperature versus pearlite growth rate plot for Fe-0.8C wt% steel. Solid lines are calculated. The data from Brown & Ridley (1969) based on particle-size analysis are regarded as the most reliable for reasons discussed in the text.

Table 1. Published values of the ferrite–cementite interfacial energy per unit area.

<table>
<thead>
<tr>
<th>reference</th>
<th>reference temperature (K)</th>
<th>method</th>
<th>$\sigma$ (J m$^{-2}$)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Das et al. (1993)</td>
<td>861</td>
<td>coarsening rate and data fitting</td>
<td>0.56</td>
</tr>
<tr>
<td>Deb &amp; Chaturvedi (1982)</td>
<td>903–963</td>
<td>coarsening rate and data fitting</td>
<td>0.248–0.417</td>
</tr>
<tr>
<td>Kramer et al. (1958)</td>
<td>1000</td>
<td>interfacial enthalpy measurement</td>
<td>0.7 ± 0.3</td>
</tr>
<tr>
<td>Martin &amp; Sellars (1970)</td>
<td>973</td>
<td>dihedral angle</td>
<td>0.52 ± 0.13</td>
</tr>
<tr>
<td>Ruda et al. (2009)</td>
<td>—</td>
<td>atomistic simulation</td>
<td>0.615</td>
</tr>
<tr>
<td>Kirchner et al. (1978)</td>
<td>—</td>
<td>interfacial enthalpy measurement</td>
<td>0.5 ± 0.36</td>
</tr>
</tbody>
</table>

(Martin & Sellars 1970), calorimetry (Kirchner et al. 1978; Kramer et al. 1958) and simulation (Ruda et al. 2009). Figure 7 compares the values of interfacial energy derived from pearlite growth-rate measurements with the independently measured values. The discrepancies are large for the lower transformation temperatures, relative to the data based on coarsening reactions and dihedral angle measurements. These are both techniques that are kinetically slow; it is possible therefore that the measured values are influenced by the segregation of solutes to the interface, which would lead to a reduction in energy. In contrast, the cementite–ferrite interfaces in pearlite are created fresh as a consequence of transformation. Entropy requires that the extent of segregation should be reduced at high temperatures.

One further difficulty is that the diffusivity $D_B$ is likely to increase with interfacial energy since a high value of the latter implies a less coherent interface. We are not able to account for this effect given the absence of relevant grain boundary-diffusion data.

Figure 8 shows a comparison of pearlite growth rates, calculated using the theory developed in this paper, against published experimental data by Frye et al. (1953) and Brown & Ridley (1969). There is a significant improvement, by a factor of between 2 and 2.5, over the growth rates calculated by Puls & Kirkaldy (1972).
A simplified theory has been proposed that combines the contributions from volume and boundary diffusivities, to represent the pearlite growth mechanism in Fe–C steels. The match with experimental data is better when compared with prior work, in spite of the fact that considerations of equilibrium at junctions between interfaces are abandoned. As might be expected, the flux through the boundary between pearlite and austenite dominates the transport of carbon at all but the highest of transformation temperatures. The theory for the first time leads to a realistic value for the activation energy for the grain boundary diffusion of carbon, less than that for volume diffusion in austenite and greater than that for volume diffusion in ferrite.

The maximum entropy and growth-rate criteria have been derived in the context of this mixed-mode diffusion theory, with the result that $S_i/S_c$ is not constant but becomes a function of the transformation temperature. The ferrite–cementite interfacial energy has been deduced assuming that the pearlite interlamellar spacing is determined by the need to maximize the entropy production rate. The energy is lower than determined in previous work, but still much higher than reported in independent experiments, possibly because the interfaces created during transformation are fresh.

It is argued that this simplified theory avoids many of the approximations required to implement a more complex model in which the shape of the transformation front is determined by equilibrium at interfacial junctions.

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Appendix A. Diffusion in ferrite

As pointed out previously, Nakajima et al. (2006) used a phase-field model to treat the possibility that a flux in the ferrite, behind the pearlite/austenite transformation front, also contributes to the growth rate of pearlite. Their model neglected the flux within the transformation front itself. We present here an analytical treatment which considers the contributions of all three fluxes, that through the volume of the austenite, through the transformation front and within the ferrite. The flux within the ferrite is supposed to lead to the thickening of cementite behind the transformation front. Equation (2.8) then becomes

$$ v S^a S^\theta \left( c^\theta - c^a \right) = 2 D_V \delta \left( c^\gamma - c^\theta \right) + \frac{12 D_B \delta \left( c^\gamma - c^\theta \right)}{S_i} - \frac{2 D_a \left( c^\alpha - c^\theta \right)}{x}, $$

where the third term on the right-hand side represents the flux within the ferrite, towards the cementite, in a region behind the transformation front; $c_i^{\alpha\gamma}$ and $c_i^{\alpha\theta}$ represent the concentrations at the respective interfaces, which are not necessarily given by equilibrium since, as will be shown below, the thickening of the cementite is interface controlled, not diffusion controlled. The average diffusion distance from the ferrite at its interface with austenite to its interface with cementite is written as $0.5xS^a$, with $x$ assumed here to be given by the ratio $v/v^\theta$; this is
because a relatively large pearlite growth rate would lead to greater diffusion distances within the ferrite. The sign of this flux is different from the other two terms because it occurs in the product phase, leading to a net reduction in the other two fluxes.

The diffusion coefficient of carbon in ferrite is known accurately (McLellan et al. 1965). The diffusivity is very large and we have calculated that the cementite growth cannot be controlled by the diffusion of carbon, but rather by interfacial mobility. Since interface-controlled growth in this case is much smaller, by three orders of magnitude, the gradient $\Delta c^\alpha$ of concentration within the ferrite will be much smaller than by using the equilibrium concentrations $c^{\alpha\gamma}$ and $c^{\alpha\theta}$.

To deal with this, the thickening rate $v^\theta$ of cementite is first calculated on the basis of the $\theta/\alpha$ interface mobility $M^{\theta\alpha}$ as

$$v^\theta = \frac{M^{\theta\alpha} \Delta G^{\theta\alpha}}{V_m},$$

where the driving force for cementite growth was calculated using MTDATA (NPL 2006). The uncertain value of mobility was taken to be $5 \times 10^{-15} \text{m}^4 \text{J s}^{-1}$ from Nakajima et al. (2006); it should strictly be a function of temperature, but was assumed constant in the absence of appropriate data.

Using the mobility equation (A2), the actual gradient within the ferrite is given by a mass balance, that the flux must equal the rate at which the cementite absorbs carbon as it grows,

$$\frac{D_\alpha (c_i^{\alpha\gamma} - c_i^{\alpha\theta})}{0.5 x S^\alpha} = v^\theta (c^{\theta\alpha} - c_i^{\alpha\theta}).$$

Figure 9 shows the results of the three-flux model against that involving only boundary and volume diffusion in the austenite. They indicate that inclusion of the flux through the ferrite would indeed lead to an increase in the growth rate, but the model without the flux within the ferrite actually represents the experimental data rather well (figure 8). The greatest uncertainty in the three-flux model arises in the mobility of the cementite–ferrite interface for which there is no experimental data. It was also pointed out earlier that evidence for the
thickening of cementite behind the transformation front is weak. To summarize, it does not at the moment seem necessary or justified to include any flux within the ferrite to explain pearlite growth data.

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