A Model of Isothermal and Non Isothermal Transformation Kinetics of Bainite in 0.5% C Steels

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The present article proposes a kinetic model for the overall transformation of austenite into bainite including both nucleation and growth of the new phase. Nucleation of bainitic ferrite laths is considered in the general framework of the classical nucleation theory. Analysis of experimental data in a low alloyed 0.5% C steel suggests that austenite grain boundary diffusion of carbon is the limiting step in the formation of ferrite nuclei. Growth of bainitic laths is controlled by volume diffusion of carbon in austenite. Nucleation and growth equations are included in a Kolmogorov–Johnson–Mehl–Avrami analysis to provide a simple relationship for the bainite fraction as a function of time and temperature which contains only two semi-empirical parameters. This model is fitted to isothermal kinetics and extrapolated to continuous cooling conditions without changing the two parameters. Comparison between predictions and experimental kinetics in two different 0.5% C steels gives a very good agreement.

KEY WORDS: phase transformation; bainite; physical model; microstructural evolution; kinetics; nucleation and growth; dilatometry.

1. Introduction

This work aims at providing a description of the transformation kinetics of austenite into bainite as a function of temperature and alloy composition. The approach adopted here explicitly takes into account the equations for nucleation and growth of the bainitic ferrite laths. In the literature, there exists a large variety of models on the kinetics of the bainite transformation, but only one attempt by Bhadeshia and co-workers is based on a physical mechanism of the transformation.1–3 The lack of physically based models for bainite kinetics is likely due to the fact that the detailed mechanisms of the bainite transformation in steels are still under debate.4–9 The aim of this work is actually not to enter this controversy but to focus on the various aspects which limit the rate of the transformation.

In the kinetic model developed by Bhadeshia, the growth of bainitic ferrite is considered to be displacive, with the carbon partitioning into the residual austenite occurring immediately after the growth of each bainite platelet. The growth is held back by the dislocation debris created by the plastic accommodation of the new platelet in the austenite matrix. The platelets thus grow to a limited size and the further progress of the transformation requires the nucleation of new platelets. The overall kinetics is then considered to be mainly controlled by the nucleation rate. Following Bhadeshia,10 the nucleation behavior of bainitic ferrite is consistent with the isothermal martensite nucleation. Accordingly, the activation energy for nucleation is taken as a linear function of the chemical Gibbs energy change involved in the formation of the nucleus. In contrast, the classical diffusional nucleation theory commonly used for ferrite would indicate that it is inversely related to the square of this quantity. The nucleation model also accounts for autocatalytic effect which gives rise to the formation of the sheaves observed experimentally. After several corrections and refinements over the years, the comparison of the model to experimental isothermal kinetics gives a reasonable agreement. But the number of fitting parameters, including the calculation of the Bs temperature, the thermodynamic limit \( T_0 \text{ or } T_2 \) (the latter accounting for the stored energy in bainitic ferrite) to the transformation and the empirical variation of the platelet size with temperature and the variation of all of them from one steel composition to another make the use of this model quite difficult as a guide to alloy design.11 All these reasons have led the present authors to revisit the problem and to propose a physically based model for the overall bainite transformation kinetics which can be extended to non-isothermal temperature treatments.

The growth aspect of the problem has been omitted in Bhadeshia’s treatment which considers instantaneous growth of the platelets and focuses on nucleation. However, in a recent experimental study of the bainite growth kinetics, the present authors showed that diffusion of carbon in austenite plays an important role in determining the rate of growth of the bainitic laths.12 This result agrees with numerous previous findings in the literature.13–15 It is worth
noting that the diffusion of interstitial atoms of carbon does not exclude any mechanism which is operative in the crystallographic transformation of the FCC lattice into the BCC lattice which might well be cooperative. Assuming para-equilibrium at the mobile interface, a simplified formulation of the Trivedi equation for diffusional growth of plate precipitates is used to calculate the bainite laths velocities. This growth model will now be implemented in an overall model for predicting the macroscopic kinetics of the bainite transformation.

The strategy adopted in the present study is the following:

- First, the overall kinetics under isothermal conditions are measured by dilatometry in an industrial alloy with a composition approximately given by Fe–0.5C–0.7Mn–0.3Cr in mass percent.
- Secondly, the nucleation kinetics will be deduced by analyzing these macroscopic kinetics in the light of the growth model presented in Ref. 12).
- Then, a classical analysis based on the idea of the “extended volume” will allow the calculation of the volume fraction of bainite from the two equations for nucleation and growth.
- After fitting the isothermal model to experimental kinetics, formulation of this model for non-isothermal kinetics will be made without changing the two fitting parameters determined in the isothermal case. Predictions for continuous cooling with varying cooling rates will be compared to experimental results.

2. Experimental

The alloys studied in this work have the compositions listed in Table 1. These alloys have been elaborated on a laboratory vacuum induction furnace under inert argon atmosphere. After hot rolling of the ingots, the material was homogenized at 1200°C for 24 h. These compositions were selected on the following grounds: alloy A is an industrial composition and will be the reference alloy of this study. Alloy B has the same carbon content as A but Mn is replaced by a high Ni content: the aim of studying this alloy is to investigate the transformation in an alloy where, due to the precipitation of cementite that accompanies bainite formation. The second assumption is that complete decomposition of the austenite is achieved during the isothermal treatment. As pointed out in the literature, bainite transformation can exhibit an “incomplete reaction phenomenon” and completion of the transformation in our specimens needs to be checked. This has been assessed by metallography and the calculation of the relative change in length expected upon complete isothermal decomposition of austenite at various temperatures which was compared to the experimental one. It is found that the transformation of austenite is indeed complete at all temperatures in the alloys investigated here. Application of Eq. (1) is then valid to calculate the isothermal kinetics.

However, during continuous cooling, the total volume fraction of bainite decreases with increasing the cooling rate and eventually becomes to zero: the microstructure then becomes fully martensitic. In our continuous cooling experiments, the final volume fraction of bainite \( X_{\text{max}} \) is measured by image analysis on metallographic sections and the fraction at a temperature \( T \) is then calculated using the following relationship:

\[
\frac{X(T)}{X_{\text{max}}} = \frac{L(T) - L_{\text{a}}(T)}{L_{\text{f}}(T) - L_{\text{a}}(T)} \quad \text{(2)}
\]

where \( L(T) \) is the actual length of the specimen, \( L_{\text{f}}(T) \) and \( L_{\text{a}}(T) \) are the calculated lengths assuming that the specimen has a fully austenitic structure at \( T \) and the volume fraction of bainite equals to \( X_{\text{max}} \) at \( T \) respectively. This method is illustrated in Fig. 1.

The conventional Avrami equation is generally used to model transformation kinetics under isothermal conditions:

\[
X(t) = 1 - \exp[-b(t-t_0)^\gamma] \quad \text{(3)}
\]

where \( t_0 \) is the effective time-lag or incubation time, \( n \) is the Avrami exponent indicative of the transformation mechanism, and \( b \) is a temperature dependent factor. An accurate determination of the Avrami exponent is required to

Table 1. Chemical composition of the alloys (in 10^{-3} mass%)

<table>
<thead>
<tr>
<th>Alloy</th>
<th>C</th>
<th>Mn</th>
<th>Ni</th>
<th>Si</th>
<th>Cr</th>
<th>Al</th>
</tr>
</thead>
<tbody>
<tr>
<td>A</td>
<td>5.29</td>
<td>688</td>
<td>-</td>
<td>285</td>
<td>32</td>
<td></td>
</tr>
<tr>
<td>B</td>
<td>478</td>
<td>&lt;1</td>
<td>4870</td>
<td>4</td>
<td>&lt;1</td>
<td>&lt;5</td>
</tr>
</tbody>
</table>
provide meaningful information concerning the transformation process. The crucial role of $t_0$ on $n$ is however seldom recognized. Here, we follow a procedure suggested by Brimacombe and co-workers. In this method, the incubation time $t_0$ is taken as an adjustable parameter which is optimized to give the best linear correlation in the Avrami plot. The advantage of this method is that the a priori knowledge of $t_0$ is not necessary while it is always difficult to identify $t_0$ from the dilatation signal which is unable to detect nucleation with the required precision.

3. Macroscopic Isothermal Kinetics: Experimental Results

Isothermal kinetics for alloy A in the temperature range 500–375°C are presented in Fig. 2. Metallographic examination of all specimens shows that the microstructure is mainly upper bainite. The best Avrami plots obtained by adjusting $t_0$ as described in the above are presented in Fig. 3. It is clear that isothermal kinetics exhibit a constant Avrami exponent close to 2 in the investigated temperature range. This result will be discussed in the following section.

4. Estimation of the Nucleation Kinetics and Nucleation Model

4.1. Summary of the Growth Model

Aiming to provide a physically based overall kinetic model of the bainitic transformation, we need a description of both the nucleation and growth kinetics of the new phase. The growth step has been treated according to Ref. In this model, the simplified version of the Trivedi model for diffusional growth of plate precipitates is used. The velocity of the growing plates is given by:

$$v = \frac{2D}{\rho} \frac{9}{16\pi} \Omega^2$$  \hspace{1cm} (4) $$

where $\Omega_*$ is an algebraic combination of the carbon supersaturation $\Omega_0$ which is calculated assuming that paraequilibrium holds at the interface:

$$\Omega_* = \frac{\Omega_0}{1 - \frac{2}{\pi} \Omega_0 - \frac{1}{2\pi} \Omega_0^2}$$  \hspace{1cm} (5) $$

According to the literature, a weighted average diffusivity of carbon in austenite is used:
Comparison between experimental data for Fe–6Ni–C alloys from Yada and Ooka \(^{15}\) and model calculations (solid lines): the effect of both the temperature and the carbon content is correctly described.

\[
D = \frac{1}{x^{1/2}} \int_{x_c}^{x} D_G^2(x) \, dx
\] ..........................(6)

where \(D_G^2(x, T)\), the composition dependant diffusivity of carbon in austenite, is provided by the empirical relationship of Kaufman et al. \(^{14}\).

Equation (4) is a combination of two unknowns, namely the velocity \(v\) and the tip radius \(\rho\). Following Zener, the classical assumption to calculate the velocity is that the system selects the tip curvature which gives a maximum growth rate. This hypothesis leads to:

\[
1/\rho = \frac{3}{32} \Omega_\star (1/\rho_c)
\] ..........................(7)

with \(\rho_c\) a critical curvature calculated by the relation:

\[
\sigma V^2/\rho_c = -\Delta G_m
\] ..........................(8)

where \(-\Delta G_m\) is the nucleation driving force of ferrite under paraequilibrium which also enters the calculation of the nucleation rate, \(V_m\) and is the molar volume of ferrite and \(\sigma\) is the ferrite–austenite interfacial energy per unit area. The latter was calculated using a linear extrapolation to low temperatures of the data compiled in references \(^{21,22}\) by Aaronson and co-workers. Such extrapolation gives a value for \(\sigma\) ranging from about 80 to 120 mJ/m\(^2\) between 500 and 375°C.

Finally, putting this particular value for in Eq. (4) gives:

\[
v_0 = \frac{27}{256\pi} \frac{D}{\rho_c} \Omega_\star^3
\] ..........................(9)

This growth model is compared to experimental results obtained by Yada and Ooka \(^{15}\) for different Fe–Ni–C steels. This comparison is shown in Fig. 4: the effect of both the temperature and the carbon content is correctly described. It is noted that the key issue is the correct prediction of both the temperature and the carbon content dependencies rather than the calculation of the absolute values which can be easily obtained by a simple multiplication constant as an additional adjustable parameter.

4.2. Deduction of the Nucleation Kinetics

An equation for the nucleation rate of laths is now required. Metallography indicates that nucleation mainly occurs on the austenite grain faces. But direct measurement of the nucleation kinetics by metallography is a difficult task and unfortunately, no convenient “in situ” method seems to be available to achieve this purpose. Qualitatively, it is however suggested that the nucleation rate decreases with the decrease in temperature. The observation of different specimens reacted at different temperatures for a short constant time \(\Delta t\) following the incubation time shows that the density of visible bainite particles in the metallographic section decreases with decreasing the temperature. In addition, it is often found that some grains have achieved complete transformation when the transformation is beginning in some others. Then the hypothesis of “saturation site”, which is often invoked when the undercooling becomes large, is not valid in the present case and nucleation is likely to proceed during the whole duration of the transformation. In the following, we propose a method to deduce the nucleation kinetics (at least its variation with temperature) from the macroscopic kinetics and a separate evaluation of the growth kinetics.

On the one hand, the general expression of Johnson–Mehl–Avrami kinetics as given in Eq. (3) can be seen as a purely phenomenological description. On the other hand, the coefficients \(b\) and \(n\) can be interpreted on physical basis. For instance, the transformation of austenite into pearlite in eutectoid steels; (b) bainitic reaction described by the linear growth of ferrite plates nucleated at austenite grain surfaces with a constant rate. The section of the plates is assumed to be essentially constant compared to the lenghtening.

Fig. 4. Schematics showing the relation between the value of the Avrami exponent and the progress of the transformation: (a) pearlitic reaction in eutectoid steels; (b) bainitic reaction described by the linear growth of ferrite plates nucleated at austenite grain surfaces with a constant rate. The section of the plates is assumed to be essentially constant compared to the lengthening.

Fig. 5. Schematics showing the relation between the value of the Avrami exponent and the progress of the transformation: (a) pearlitic reaction in eutectoid steels; (b) bainitic reaction described by the linear growth of ferrite plates nucleated at austenite grain surfaces with a constant rate. The section of the plates is assumed to be essentially constant compared to the lengthening.
overall volume fraction\(^{23}\):

\[
X(t) = 1 - \exp\left(-\frac{\pi}{3} N v^3 t\right) \tag{10}
\]

Here, \(n\) is equal to 4, which accounts for the three-dimensional linear growth of the spheres and a constant nucleation rate. The coefficient \(b\) is a product of a geometrical constant, the nucleation rate and the cube of the growth rate. Christian\(^{21}\) has compiled the expressions of \(b\) and \(n\) in other various simple situations. Investigating more complex transformations starting at grain boundaries, others models such as those developed by Johnson and Mehler\(^{24}\) and later by Cahn\(^{25}\) should be considered. The formulation of the volume fraction of the product phase slightly deviates from the pure Avrami law (Eq. (3)). Nevertheless when the isothermal kinetics is still satisfactorily described by this simple law, the interpretation of \(n\) and \(b\) as in the above case is still valid. Then, if the growth rate \(v\) is measured independently, the analysis of \(b\) can provide an estimation of the nucleation rate \(N\).

In the case of bainitic transformation, assuming a constant growth rate \(v\) (as observed experimentally) and a constant nucleation rate would lead to \(n=2\). Growth is indeed considered to occur mainly in one dimension assuming that thickening of the plates can be neglected compared to their lengthening. This picture of the transformation is schematized in Fig. 5(b). This is exactly the value suggested by the previous Avrami analysis for which \(n\) was considered as a purely adjustable parameter. Now forcing \(n=2\) in the Avrami analysis of the experimental global kinetics will allow us to a simple interpretation for \(b\), now labeled \(b_2\). The coefficient \(b_2\) is proportional to the product \(N v_0\). Therefore, the ratio \(b_2/v\) is directly proportional to the nucleation rate. Since we know that the temperature dependence of \(v\) is accurately described by the Trivedi formula (see Fig. 4 and also Ref. 12)), the ratio \(b_2/v\) provides an estimate of the nucleation rate \(N\) for different temperatures. Change of this ratio with temperature for alloy A is presented in Fig. 6.

The salient feature of this figure is that the nucleation rate decreases with decreasing temperature. This result excludes a displacive mechanism for nucleation: a diffusionless mechanism for the formation of the nuclei would involve only the driving force for the transformation without composition change which increases with decreasing the temperature. The nucleation rate would then increase as the temperature is lowered.\(^{26,27}\) This is in contradiction to what is observed experimentally. The nucleation step is therefore a diffusional process and since the nucleation rate decreases with decreasing temperature, it suggests that the temperature dependence is mainly controlled by the diffusion coefficient and not by the driving force. The hypothesis of a diffusional mechanism for the formation of a nucleus is also supported by the linear correlation pointed out by Bhadeshia\(^{3}\) between the apparent Bs temperature and the driving force for diffusional nucleation of ferrite, which allows carbon partition between the nucleus and the matrix. It is worth noting that this correlation is not as satisfactory when the driving force for martensitic nucleation—or any other driving force—is considered. This empirical result suggests that carbon redistribution plays an important role during the formation of a nucleus.

4.3. Nucleation Model

The general framework of classical nucleation theory is then used and the nucleation rate is written:

\[
\dot{N} = K \exp\left(-\frac{Q_D}{RT}\right) \exp\left(-\frac{\Delta G^*}{RT}\right) \tag{11}
\]

where the successive exponential terms account for the diffusional process limiting the formation of the nucleus and the work required to form this nucleus respectively; \(K\) being a constant. As it is pictured in Fig. 7, the diffusional exponential term has the strongest influence in the low temperature range where the driving force is large and which corresponds to the bainite transformation.

Evaluation of the work \(\Delta G^*\) required to form a critical nucleus can be carried out in the present case according to the “pill box” nucleation model developed by Aaronson and co-workers\(^{28}\):

\[
\Delta G^* = \frac{4\pi V_m^2 \sigma^2 e}{\Delta G_M^2} \tag{12}
\]

where \(V_m\) is the molar volume of ferrite, \(\epsilon=7.5\) mJ/m\(^2\) and \(\Delta G_M\) is the Gibbs energy change accompanying the formation of a nucleus under paraequilibrium conditions.
(carbon redistribution only). According to the data compiled in references, the interfacial energy \( \sigma \) is estimated to vary from about 80 to 120 mJ/m\(^2\) between 500 and 375°C. Then, calculation shows that \( \Delta G^* \) varies from 1.7\( \text{kT} \) to 1.1\( \text{kT} \) between 500 and 375°C respectively. The value of \( \exp(-\Delta G^*/kT) \) changes from 0.18 to 0.33 in the same temperature range. This temperature dependence remains low compared to the carbon diffusivity in austenite which will change by two orders of magnitude between these two temperatures. The temperature dependence of the nucleation rate is then mainly controlled by the diffusion coefficient. Now the temperature dependence of the second exponential term in Eq. (11) can be neglected. A classical Arrhenius analysis of the data in Fig. 6 gives a value of about 47 kJ/mol for the activation energy \( Q_D \). The activation energy for bulk diffusion of carbon in austenite is given by Kaufman et al.,

\[
Q_C = 38300 - 1.9 \times 10^3 x_C^6 + 5.5 \times 10^5 x_C^{0.2} \quad \text{(13)}
\]

with \( Q_C \) in cal/mol and \( x_C \) the molar carbon concentration. For an initial concentration of 0.5% in mass, a value of 140 kJ/mol is expected which is three times higher than the above experimental value. In other words, the nucleation rate is less sensitive to the temperature than the bulk diffusivity of carbon in austenite. The value of the ratio \( Q_C/Q_D \) found here is considered to be typical for a grain boundary diffusion process. It is then proposed that the nucleation rate of bainite laths is likely to be controlled by austenite grain boundary diffusion of carbon.

5. Isothermal Global Kinetics

The next part of this study is aimed at providing a relationship between bainite fraction, time, temperature and alloy composition with a limited number of adjustable parameters. This will be done on the basis of the equations developed in the above for nucleation and growth of the bainite phase. First, the model will be formulated for isothermal conditions and the value of the parameters will be determined by fitting the model to isothermal kinetics in alloys A and B. Secondly, a formulation of the model under continuous cooling condition will be derived and predictions for various cooling rates, keeping the parameters values unchanged, will be compared to experimental kinetics for alloy A.

Because of the crystallographic character of the transformation, the growth of bainite is highly anisotropic and the geometrical description of the transformation at the scale of the austenite grain remains difficult. In order to keep the model simple, the general equation due to Kolmogorov–Johnson–Mehl–Avrami is employed:

\[
X(t) = 1 - \exp(-Ve/V_0) \quad \text{(14)}
\]

where \( Ve \), usually called the “extended” volume, represents the volume that would be filled by the grains of the new phase without impingement and \( V_0 \) the initial volume to be transformed. Under isothermal conditions, the “extended” volume is given by:

\[
Ve(t) = \int_0^t (\text{number of laths nucleated between } t \text{ and } \tau + d\tau) \times (\text{volume at } t \text{ of a lath nucleated at } \tau) \times d\tau
\]

The integration gives:

\[
Ve(t) = \alpha N v_0 (t-t_0)^2 \quad \text{(15)}
\]

At this stage, there is only one parameter \( K \), which has to be adjusted from experimental isothermal kinetics. The second parameter \( K_i \) is the ratio accounting for austenite grain boundary diffusion during nucleation and its value has been previously determined as 1/3.

The comparison between experimental and predicted isothermal kinetics for both alloys A and B is presented in Fig. 8 and shows a satisfactory agreement. It is worth noting that the value of the parameter \( K_i \) which indicates a deviation from the maximum growth rate calculated by the Trivedi equation is lower (indicating a larger deviation) in the high alloyed steel (alloy B) than in the low alloyed one (alloy A). It is likely that Ni atoms interact with the moving interface thus giving rise to a slowing down effect.

Fig. 8. Comparison between experimental and calculated isothermal kinetics for alloys A and B. Values of the constants are: \( K_i=0.33 \) in the two cases; \( K_i=0.285 \) and 0.140 for alloys A and B respectively.
Incubation time $t_0$ in the Eq. (17) is obtained experimentally or may be satisfactorily modeled in terms of a Russel’s type theory.\textsuperscript{34,35) Namely, incubation times are of the general form:

$$t_0 < \frac{T}{D \Delta G_m^2} \quad \text{(18)}$$

where $D$ is a diffusion coefficient characteristic of the nucleation mechanism. Accordingly, we can write:

$$D = \exp(-K_2 Q_g / RT) \quad \text{(19)}$$

Therefore, the incubation time can be calculated using the following equation which requires only one additional adjustable constant $K_3$:

$$t_0 = K_3 \frac{T \cdot \exp(K_2 Q_g / RT)}{v_c \Delta G_m} \quad \text{(20)}$$

This equation has been compared to the experimental incubation time deduced from the Avrami analysis explained in details in part 2 of this paper and a good agreement was found.\textsuperscript{36)}

6. Non-isothermal Global Kinetics

Continuous cooling treatments are of principal interest in industrial schedules. The KJMA analysis explained in the above can be easily extended to such conditions. The “extended” volume of bainite at a temperature $T_1$ along the cooling path is simply given by:

$$V_e(T_1) = \int_{T_1}^{T} \left( \text{number of laths nucleated between } T \text{ and } T + dT \times \text{(volume at } T_1 \text{ of a lath nucleated at } T \right) \times dT \quad \text{(21)}$$

where Bs is the critical temperature below which bainite starts to form. Integration of Eq. (21) gives:

$$\frac{Ve(T_1)}{V_c} = \frac{1}{T^2} \int_{T_1}^{T} N(T) \left( \int_{T}^{T_1} v(T) dT \right) dT \quad \text{(22)}$$

where $T$ is the constant cooling rate. It is then possible to calculate the fraction of bainite as:

$$X(T_1) = 1 - \exp \left[ -\frac{K_1}{T^2} \int_{T_1}^{T} \exp(-K_2 Q_g / RT) \zeta(T, T_1) dT \right] \quad \text{(23)}$$

with $\zeta(T, T_1) = \int_{T_1}^{T} v(T) dT$. Again this equation involves only two parameters $K_1$ and $K_2$ which, according to the physical content of the model, should remain unchanged in both conditions of thermal treatment. The predicted continuous cooling kinetics for alloy A are presented in Fig. 9. A good qualitative tendency is obtained: a low cooling rate allows the transformation to go to completion while the final bainite fraction diminishes as the cooling rate is increased. It will eventually become to zero, providing a prediction for the critical quenching rate above which a fully martensitic microstructure is obtained.

Comparison with experimental results should be carried
out with care. It is necessary to prevent ferrite and pearlite formation at high temperature since the presence of these components would noticeably modify the nucleation kinetics of bainite. Avoiding the formation of ferrite and pearlite during cooling has been achieved by rapid cooling of the specimens at 80°C/s from the austenitisation temperature to a temperature of 550°C which is slightly above the Bs temperature deduced from isothermal experiments in alloy A. The kinetics of the bainite transformation during cooling from this temperature are observed. Such temperature treatments are pictured in Fig. 10(a). The correspondent kinetics are given in Fig. 10(b) and compared to the predictions of the model. A good quantitative agreement is found at slow cooling rates for both the transformation rate and the final bainite fraction. However, at higher cooling rates, the model overestimates the maximum fraction of bainite obtained after cooling. The reason for this discrepancy is not straightforward. It might come from a transition from upper bainite, with the presence of carbides at the austenite–ferrite interface acting as carbon sinks, to lower bainite where carbides precipitate inside the ferrite and cannot absorb the carbon rejected at the lath tips. Accordingly, the transformation kinetics of lower bainite are expected to be slower than the extrapolated transformation kinetics of upper bainite at the same temperature, and the final volume fraction of bainite is overestimated by the model.

8. Conclusion

Relying on a model for the growth kinetics of bainite laths based on valuable experimental data and careful evaluation of the thermodynamic properties in the bainite temperature range, we have developed a kinetic model for the overall transformation. The temperature dependence of the nucleation rate has been derived by combining the experimental global isothermal kinetics and the calculated growth rate. It has been shown that the nucleation rate of bainite decreases with decreasing temperatures. This result cannot be explained by a displacive mechanism which would predict just the opposite effect. The classical approach for nucleation is adopted here. As the driving force is large in the bainite temperature range and the nucleus will form with a very low carbon content, it is likely that the temperature dependence of the nucleation rate is only limited by carbon diffusion in austenite. Analysis of the experimental data in a low alloyed 0.5% C steel suggests that austenite grain boundary diffusion of carbon is the limiting step. Then, nucleation and growth equations are combined in a Kolmogorov–Johnson–Mehl–Avrami analysis: it provides a simple relationship for the bainite fraction which contains only two semi-empirical parameters. This relation is written for both isothermal and continuous cooling conditions. Because of their physical nature, the two adjustable parameters remains unchanged in both cases. This general model has been successfully compared to experimental kinetics in two different 0.5% C steels and can be extended to other carbon concentrations.

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