Modelling of isothermal formation of pearlite and subsequent reaustenitisation in eutectoid steel during continuous heating

F. G. Caballero, C. Capdevila and C. García de Andrés

Dr. F.G. Caballero and Dr. C. García de Andrés are in the Department of Physical Metallurgy, Centro Nacional de Investigaciones Metalúrgicas (CENIM), CSIC, Avda. Gregorio del Amo, 8, 28040 Madrid, Spain. Dr. C. Capdevila, is in the Department of Materials Science and Metallurgy, University of Cambridge, Pembroke Street, Cambridge CB2 3QZ, UK

Three different morphologies of pearlite have been formed isothermally at three different temperatures in a eutectoid steel. Moreover, the interlamellar spacing of pearlite was calculated using Zener and Hillert theoretical method. Experimental results suggest that the growth of pearlite is mainly controlled by volume diffusion of carbon in austenite in the temperature range studied in this steel. In addition, a model that describes pearlite-to-austenite transformation during continuous heating in a eutectoid steel has been developed. The influence of structure parameters, such as interlamellar spacing and edge length of pearlite colonies, on the transformation kinetics has been experimentally studied and considered in the modeling. It has been found that the kinetics of pearlite-to-austenite transformation is slower the coarser the initial pearlite microstructure. Experimental validation of this model has been carried out and a good agreement (accuracy higher than 90% in square correlation factor) between experimental and calculated values has been found.

1 Introduction

Pearlite is a lamellar product of eutectoid decomposition, which can be formed in steels and non-ferrous alloys during transformations under isothermal or continuous cooling.^{1,2} A pearlite nodule is composed of multiple colonies; each colony has parallel lamellae, which are orientated differently with respect to lamellae in adjacent colonies. This also exhibits a wide range of interlamellar spacing in different colonies because of the intersection of pearlite colonies at different angles with the polishing plane. The interlamellar spacing is reflected by the diffusion kinetics at the transformation front and is a sensitive parameter which, in a particular steel, is larger as the transformation temperature increases.³ Mehl and co-workers³ demonstrated that the spacing decreases as the degree of undercooling, ΔT , below the eutectoid temperature, increases. Zener⁴ provided the first theoretical analysis of these observations, which allows **a calculation of** the interlamellar spacing of pearlite as a function of undercooling.

Pearlite transformation in steels is reconstructive and known to show a constant growth rate.⁵ The growth rate of pearlite is **generally** believed to be controlled by either volume diffusion of carbon^{4,6} or by boundary diffusion of substitutional alloying elements.⁷ **There are some studies debating the importance of boundary diffusion of carbon, especially at intermediate transformation temperatures⁸, although this possibility is not consider in the present work.**

In **this paper**, three different morphologies of pearlite were formed isothermally at three different temperatures in a eutectoid steel. As Mehl et al. and Zener reported^{3,4}, the pearlite is finer the lower formation temperature. Moreover, the interlamellar

spacing was calculated using the theoretical method proposed by Zener and Hillert.^{4,6,7} Experimental results suggest that the growth of pearlite is mainly controlled by volume diffusion of carbon in austenite in the temperature range studied in this steel.

On the other hand, the formation of austenite during heating differs in many ways from those transformations that occur during cooling of austenite. In this sense, the kinetics of transformation of austenite during cooling can be described completely in terms of chemical composition and grain size of **only** this phase. However, the microstructure from which austenite may grow during heating can be **much more complex**. Therefore, to describe the kinetics of austenite formation additional variables are needed. Factors such as particle size, the distribution and chemistry of individual phases, homogeneity, the presence of non-metallic inclusions, all are important.⁹⁻¹² In the case of reaustenitization from pearlite the most relevant structural factor to be considered is the interlamellar spacing of pearlite.¹³

In fully pearlitic steel, austenite nucleates heterogeneously at the junctions between pearlite colonies. This is in spite of the relatively large amount of interlamellar surfaces available within the pearlite colonies, which seem to be much less effective as sites for the nucleation of austenite.¹⁴ The rate of growth of the austenite is primarily controlled by the rate of carbon diffusion in the austenite between adjacent pearlitic cementite lamellae, but may also be influenced by boundary diffusion of substitutional alloying elements at low temperature.¹⁰

Models of specific metallurgical approaches exist for isothermal reaustenitization from different initial microstructures.^{10,14-20} However, none of these is likely to be applicable to non-isothermal conditions. In this work, a model is also presented for

the reaustenitization from pearlite during continuous heating in a eutectoid steel. The influence of **parameters such as the interlamellar spacing of pearlite and the length of the edge of the pearlite colonies** on transformation kinetics has been considered in the model. Results of modeling have been experimentally validated at the three different morphologies of pearlite previously formed isothermally.

2 Experimental Procedure

2.1 ISOTHERMAL FORMATION AND MORPHOLOGICAL CHARACTERIZATION OF PEARLITE

A eutectoid carbon steel with an actual composition (wt-%) of 0.76C, 0.24Si, 0.91Mn and 0.013P was used. The following heat treatments were carried out to yield in this steel fully pearlitic microstructures with different morphologies. Specimens were austenitized for 5 min at 1273 K, isothermally transformed at three different temperatures and subsequently cooled rapidly to room temperature. Table 1 lists all the temperatures and holding times used for the isothermal formation of pearlite with different morphological parameters in this steel.

The heating and cooling devices of an Adamel Lhomargy DT1000 high-resolution dilatometer have been used to perform all the heat treatments previously mentioned. This dilatometer is equipped with a radiation furnace for heating. The power radiated by two tungsten filament lamps is focused on the specimen by means of a bi-elliptical reflector. The temperature is measured with a 0.1 mm diameter Chromel-Alumel (type K) thermocouple welded to the specimen. Cooling is carried out by blowing a jet of

helium gas directly onto the specimen surface. The helium flow rate during cooling is controlled by a proportional servovalve. The excellent efficiency of heat transmission and the very low thermal inertia of the system ensure that the heating and cooling rates, ranging from 0.003 K/s to approximately 200 K/s, remain constant.

Specimens were polished in the usual way and finished on 0.25 μm diamond paste for metallographic examination. It was found that long polishing times, particularly on the 6 μm pad, resulted in a deep worked layer that produced a distorted lamellar structure. Cementite plates in final microstructure seem to bend and eventually fracture during plastic deformation of ferrite plates induced by long polishing periods. Polishing times up to five minutes on the 0.25 μm pad were found to be adequate to remove evidence of surface deformation caused by previous pads. Deep primary etching with a solution of picric acid in isopropyl alcohol with several drops of Vilella's reagent was used to ensure that any deformed layer introduced by polishing was removed. This etching was eliminated using the 1 and 0.25 μm diamond pads with almost no pressure being exerted on the sample for no longer than 3 to 4 min. The sample was then etched again, this time lightly, and polished carefully on the 1 and 0.25 μm diamond pad. Finally, a light etch was given to the sample.

This repeated polishing-etching procedure allows **disclosure of** the morphology of pearlite in specimens MORF1 and MORF2 on a JEOL JXA - 820 scanning electron microscope (SEM) (Figs. 1(a) and (b)). The morphology of pearlite in specimen MORF3 was revealed by transmission electron microscopy. Cylindrical samples were sliced into 100 μm thick discs and subsequently ground down to a thickness of 50 μm on wet 800 grit silicon carbide paper. These foils were finally electropolished

until perforation occurred in a twin-jet electropolishing unit at room temperature and a voltage of 100 V using a solution of 5% perchloric acid, 15% glycerol and 80% methanol. Thin foils were examined in a JEOL – 200CX transmission electron microscope (TEM) at an operating voltage of 200 kV (Fig. 1(c)).

Two parameters, the mean true interlamellar spacing (σ_o) and the area per unit volume of the pearlite colonies interface (S_v^{PP}), characterize the morphology of pearlite.¹⁵ The values of σ_o were derived from electron micrographs according to Underwood's intersection procedure. Underwood²¹ recommends determining the mean random spacing σ_r first to estimate the mean true spacing σ_o . For this purpose, a circular test grid of diameter d_c is superimposed on an electron micrograph. The number n of intersections of lamellae of carbide with the test grid is counted. This procedure is repeated on a number of fields chosen randomly. Then, the mean random spacing σ_r is calculated from:

$$\sigma_r = \frac{\pi d_c}{nM} \quad (1)$$

where M is the magnification of the micrograph.

Saltykov²² has shown that, for pearlite with a constant spacing within each colony, the mean true spacing σ_o is related to the mean random spacing σ_r by:

$$\sigma_o = \frac{\sigma_r}{2} \quad (2)$$

The values of S_v^{PP} were measured by counting the number of intersections (n') of the pearlite colony boundaries with a circular test grid of diameter d'_c as reported by Roosz et al.¹⁵ Then, the area per unit volume of the pearlite colonies interface is:

$$S_v^{PP} = \frac{2n'M}{\pi d'_c} \quad (3)$$

Approximating the pearlite colony by a truncated octahedron, the edge length of the pearlite colonies (a^P) is calculated from the area per unit volume S_v^{PP} with the following expression²³:

$$S_v^{PP} = \frac{6(1+2\sqrt{3})}{8\sqrt{2}(a^P)^3} (a^P)^2 = \frac{3(1+2\sqrt{3})}{4\sqrt{2}a^P} \quad (4)$$

Data for σ_o , S_v^{PP} and a^P are listed in Table 2.

2.2 DILATOMETRIC AND METALLOGRAPHIC ANALYSIS OF REAUSTENITIZATION FROM PEARLITE

The experimental validation of the reaustenitization model developed in this work was carried out using the dilatometer **described above**. To analyze the progress of pearlite-to-austenite transformation, interrupted heating experiments were carried out by quenching. Dilatometric specimens with the three different morphologies of pearlite (see Table 2) were heated at a constant rate of 5 K/s. Each test was repeated

three times. Heating dilatometric curves were analyzed to determine the start temperature (Ac_1) and the end temperature (Ac_3) of pearlite-to-austenite transformation and then several quench-out temperatures were selected in order to investigate the progress of the transformation. Figure 2 shows the seven selected quench-out temperatures on a dilatometric curve. They were defined as follows: $T_a=Ac_1-5$ K, $T_b=Ac_1$, T_c , T_d , and T_e , are the temperatures at the maximum, inflexion point and minimum, respectively, of the heating dilatometric curve, $T_f=Ac_3$ and $T_g=Ac_3+10$ K. The Ac_1 and Ac_3 temperatures are the average value of three similar tests. All these temperatures at which heating was interrupted by quenching are listed in Table 3.

Austenite, which is formed inside pearlite, transforms to martensite during quenching. Thus, the progress of reaustenitization process is determined throughout the evolution of the volume fraction of martensite. Specimens from interrupted heating experiments were polished in the usual way for metallographic examination. Le Pera's reagent²⁴ was used to reveal martensite formed during quenching. The quantitative measurement of martensite volume fraction was determined by point counting²¹.

3 Results and Discussion

3.1 CALCULATION OF THE INTERLAMELLAR SPACING OF ISOTHERMALLY FORMED PEARLITE IN A EUTECTOID STEEL

The growth rate of pearlite is believed to be controlled by either volume diffusion of carbon^{4,6} or by boundary diffusion of substitutional alloying elements.⁷ When the growth rate of pearlite is controlled by the bulk diffusion of atoms in austenite ahead of the interface, the diffusion of carbon may play a more important role than that of substitutional alloying elements. Since the diffusivity of the substitutional alloying elements in austenite is far smaller than that of carbon, the substitutional alloying elements may not diffuse a long distance during the reaction. In that case, Zener⁴ proposed the following relationship between the interlamellar spacing, σ_o , and the theoretical critical spacing at zero growth rate, σ_c , based **on** the maximum growth rate criterion. That leads to an expression for σ_o as a function of the pearlite formation temperature:

$$\sigma_o = 2\sigma_c = \frac{4T_e\sigma_{\alpha\theta}}{\Delta H_v(T_e - T)} \quad (5)$$

where T_e is the eutectoid temperature, T is the formation temperature, $\sigma_{\alpha\theta}$ is the interfacial energy per unit area of the ferrite-cementite lamellar boundary in pearlite and ΔH_v is the change in enthalpy of transformation per unit volume.

Puls and Kirkaldy²⁵ reported the following $\sigma_{\alpha\theta}$ and ΔH_v values, the former is **an** experimental value due to Kramer et al.²⁶

$$\sigma_{\alpha\theta} / (J / m^2) = 1 \quad (6)$$

$$\Delta H_v / (J / m^3) = 6.07 \times 10^8 \quad (7)$$

From Eqs. (6) and (7), and considering an approximated eutectoid temperature, T_e , of 1000 K, Eq. (5) could be rewritten as follows:

$$\sigma_o / \mu m = \frac{6.5898}{(1000 - T)} \quad (8)$$

with T given in K.

When the partitioning of the substitutional alloying elements is substantial during the growth of pearlite, boundary diffusion of the alloying elements may control the growth rate of pearlite. In that case, the maximum growth rate criterion of Zener⁴ gives an expression for σ_o as a function of the pearlite formation temperature as follows:

$$\sigma_o = \frac{3}{2} \sigma_c = \frac{3T_e \sigma_{\alpha\theta}}{\Delta H_v (T_e - T)} \quad (9)$$

Substituting Eqs. (6) and (7) in Eq. (9) and considering again an approximated eutectoid temperature, T_e , of 1000 K, an expression of the interlamellar spacing as a function of the formation temperature is obtained in this steel for the case of boundary diffusion of controlled pearlite growth with T given in K:

$$\sigma_o / \mu m = \frac{4.9423}{(1000 - T)} \quad (10)$$

Figure 3 represents the variation of the interlamellar spacing as a function of undercooling **for rate controlling mechanism of pearlite growth via volume diffusion of carbon and via boundary diffusion of substitutional alloying elements**, calculated according to Eqs. (8) and (10), respectively. Moreover, this figure shows the experimental values of interlamellar spacing listed in Table 2. This figure suggests that the growth of pearlite is mainly controlled by volume diffusion of carbon in austenite in the temperature range studied and confirms the applicability of Zener and Hillert theory to this particular steel.

3.2 MODELING OF NON-ISOTHERMAL PEARLITE-TO-AUSTENITE TRANSFORMATION KINETICS

Nucleation and growth processes under isothermal condition can be described in general using the Avrami's equation²⁷:

$$x = 1 - \exp(-Kt^n) \quad (11)$$

where x represents the formed austenite volume fraction, K is a constant for a given temperature, t is the time, and n is a constant characterizing the kinetics. Roos et al.¹⁵ obtained a value of $n=4$ from their measured data under isothermal condition. According to Christian⁵, with a spherical configuration, a value of $n=4$ means that the nucleation rate ($\dot{N}$) and the growth rate (G) are constant in time. This gives a transformed volume fraction of:

$$x = 1 - \exp\left(-\frac{\pi}{3} \dot{N} G^3 t^4\right) \quad (12)$$

Roosz et al.¹⁵ proposed a temperature and structure dependence of $\dot{N}$ and G as a function of the reciprocal value of overheating ($\Delta T = T - A_{c1}$) as follows:

$$\dot{N} = f_N \exp\left(\frac{-Q_N}{k\Delta T}\right) \quad (13)$$

$$G = f_G \exp\left(\frac{-Q_G}{k\Delta T}\right) \quad (14)$$

where Q_N and Q_G were denominated activation energies of nucleation and growth by Roosz et al.¹⁵, respectively, k is Boltzmann's constant, and f_N and f_G are the functions representing the influence of the structure on the nucleation and growth rates, respectively. The activation energies of nucleation and growth, Q_N and Q_G , were determined experimentally by Roosz et al.¹⁵ in a eutectoid steel ($Q_N=3.5\times 10^{-22}$ J and $Q_G=4.1\times 10^{-22}$ J).

The three different ferrite/cementite interface sites for austenite nucleation in fully pearlitic steels are: A) the interface of ferrite and cementite platelets, B) the surfaces of the pearlite colony, and C) the triple points of the pearlite colony intersections. Several authors¹⁴⁻¹⁶ reported that the nucleation of austenite inside pearlite takes place preferentially at the points of intersection of cementite with the edges of the pearlite colony (type C).

Approximating the pearlite colony as a truncated octahedron, the number of nucleation sites per unit volume (type C) is calculated as follows:

$$N_C = \frac{\left(\frac{36}{3} \frac{a^P}{\sigma_o} \right)}{8\sqrt{2}(a^P)^3} = \frac{36a^P}{24\sqrt{2}(a^P)^3 \sigma_o} \approx \frac{1}{(a^P)^2 \sigma_o} \quad (15)$$

where a^P is the edge length of the pearlite colony and σ_o is the interlamellar spacing. Bearing in mind that the rate of nucleation increases as the pearlite interlamellar spacing decreases and the edge length of the pearlite colony increases²⁸, the function f_N in Eq. (13) would have the general form:

$$f_N = K_N \frac{(a^P)^n}{\sigma_o^m} (N_C)^i \quad (16)$$

where K_N , n , m , and i are empirical parameters. These parameters were adjusted in order to obtain good fit between theory and the experimental austenite volume fraction curves. Different cases of structure dependence for n , m , and i were investigated and it was found out that the measured values of austenite volume fraction as a function of temperature can be best described with $n=6$, $m=1$, and $i=\frac{5}{3}$.

Assuming that the growth of austenite into pearlite is controlling by volume diffusion of carbon in austenite¹⁵, the function f_G in Eq. (14) representing the structure dependence on the growth rate can be expressed as follows,

$$f_G = K_G \frac{1}{\sigma_o} \quad (17)$$

where K_G is an empirical constant.

Substituting Eqs. (16) and (17) in Eqs. (13) and (14), respectively:

$$\dot{N}/(1/mm^3s) = 2.2 \times 10^{12} \frac{(a^P)^6}{\sigma_o} (N_C)^{\frac{5}{3}} \exp\left(\frac{-25.38}{\Delta T}\right) \quad (18)$$

$$G/(mm/s) = 10^{-10} \frac{1}{\sigma_o} \exp\left(\frac{-29.7}{\Delta T}\right) \quad (19)$$

K_N and K_G empirical constants were adjusted from the experimental austenite volume fraction results.

The difficulties in treating non-isothermal reactions are **mainly** due to the independent variations of growth and nucleation rates with temperature, described in Eqs. (18) and (19). The problem is only undertaken when the rate of transformation depends exclusively on the state of the assembly and not on the thermal path by which the state is reached.²⁷ Reactions of this type are called isokinetic. Avrami defined an isokinetic reaction by the condition that the nucleation and growth rates are proportional to each other (i.e. they have the same temperature variation). This leads to the concept of additivity and Scheil's rule.²⁹

Since Avrami's condition for an isokinetic reaction is not here satisfied, a general equation to describe the non-isothermal overall pearlite-to-austenite transformation in pearlitic steel was derived integrating the Avrami's equation over the whole

temperature range where the transformation takes place.³⁰ In this sense, we have taken logarithms in Eq. (11) and then it was differentiated,

$$d\left(\ln \frac{1}{1-x}\right) = \frac{dx}{1-x} = \frac{4\pi}{3} \dot{N} G^3 t^3 dt \quad (20)$$

If we consider a constant rate for the heating condition, time can be expressed as follows:

$$dt = \frac{dT}{\dot{T}} \quad t = \frac{\Delta T}{\dot{T}} \quad (21)$$

and substituting into Eq. (20):

$$\frac{dx}{1-x} = \frac{4\pi}{3} \dot{N} G^3 \frac{\Delta T^3}{\dot{T}^4} dT \quad (22)$$

Integrating in Eq. (22) in $[0, x]$ and $[Ac_1, T]$ intervals on the left and on the right sides, respectively, it can be concluded that:

$$x = 1 - \exp\left(- \int_{Ac_1}^T \frac{4\pi}{3 \dot{T}^4} \dot{N} G^3 \Delta T^3 dT\right) \quad (23)$$

where $\dot{T}$ is the heating rate and, $\dot{N}$ and G are given by Eqs. (18) and (19). The eutectoid temperature A_{c1} of the steel was calculated using Andrews' formula.³¹

3.3 EXPERIMENTAL VALIDATION OF THE MODEL

Figure 4 shows the experimental and calculated austenite formation kinetics plotted as a function of temperature for three different morphologies of pearlite. R^2 is the square correlation factor of the experimental and calculated volume fraction of austenite formed at different temperatures. This parameter quantifies the accuracy of the model. The figure suggests that austenite transformation starts later and appears to be slower the coarser the initial pearlite microstructure. Experimental results for the austenite volume fraction are in good agreement with the predicted values from the model proposed in this work. The accuracy of this model is in the three cases higher than 90% which can be considered excellent for a kinetics model bearing in mind the experimental difficulties for its validation.

Figure 5 shows microscopic evidences of how the pearlite-to-austenite reaction occurs for the three morphologies of pearlite throughout micrographs from interrupted heating samples at the start, intermediate and finish stages of the reaustenitization process. Le Pera's reagent²⁴ reveals pearlite as a darker phase in the microstructure, whereas martensite formed during quenching appears as lighter regions in the micrographs. Microstructures in Figs. 5(a), (d) and (g) are formed mainly of pearlite and some grains of martensite. At these quench-out temperatures, the pearlite-to-austenite transformation starts, and those temperatures are considered the transformation start temperature, T_s , of the three different morphologies of

pearlite. Figures 5(b), (e) and (h) show an intermediate stage of the reaction. Finally, Figs. 5(c), (f) and (i) represent microstructures formed mainly of martensite. At this stage, the transformation pearlite-to-austenite has reached completion and those quench-out temperatures are considered the transformation finish temperature, T_F , of the three different morphologies of pearlite.

Comparison between experimental and calculated results of start (T_S) and finish (T_F) temperatures of pearlite-to-austenite transformation is shown in Figs. 6(a) and 6(b), respectively. T_S temperature is considered to be the temperature at which a volume fraction of austenite of approximately 0.01 is formed during continuous heating. Moreover, T_F temperature has been defined as the temperature at which a volume fraction of austenite of approximately 0.99 is present in the microstructure. Points lying on the line of unit slope show a perfect agreement between experimental and calculated values. The accuracy of the calculations is quantified by R^2 . Figure 6(a) shows a difference of no more than 5 K between the experimental and calculated T_S values. That difference can be a reflection of the fact that this temperature is difficult to determine experimentally by optical microscopy on interrupted heating specimens at previously selected quench-out temperatures on dilatometric curves. Likewise, Fig. 6(b) shows a very good agreement (R^2 of 0.9992) between experimental and predicted T_F values. Furthermore, Figs. 6(a) and (b) suggest that both T_S and T_F temperatures are higher the **coarser** the initial pearlite. The morphology of initial pearlite is therefore a fundamental variable in the inevitable austenitization process during the heat treatment of steels.

4 Conclusions

1. Three different morphologies of pearlite were isothermally formed at three temperatures in a eutectoid steel (0.76C-0.91Mn-0.24Si-0.013P). The interlamellar spacing of these morphologies were measured by electron microscopy. Moreover, the interlamellar spacing of pearlite was calculated as a function of the undercooling from Zener and Hillert theoretical model. It has been demonstrated that for this steel, the growth of pearlite is controlled by the diffusion of carbon in austenite in the studied temperature range.
2. Since conditions to apply Scheil's rule are rarely satisfied, a mathematical model applying the Avrami's equation has been used to reproduce the kinetics of the pearlite-to-austenite transformation in a eutectoid steel during continuous heating. The model proposes two functions, f_N and f_G , which represent the dependence of nucleation and growth rates, respectively, on the structure. The influence of the parameters that characterize the morphology of pearlite on the overall transformation kinetics **has also been** experimentally studied. It has been found that austenite formation commences at higher temperature the coarser the pearlite. Moreover, the kinetics of the transformation **are** slower the coarser the initial pearlite microstructure. Experimental results for the austenite volume fraction and start and finish temperatures of pearlite-to-austenite transformation are in good agreement with the predicted values from the model proposed in this work (accuracy higher than 90% in square correlation factor).

5 Acknowledgements

The authors acknowledge financial support from Consejería de Educación y Cultura de la Comunidad Autónoma de Madrid (CAM 07N/0065/1998).

6 References

1. N. RIDLEY: *Metall. Trans.*, 1984, **16A**, 1019-1036.
2. N. RIDLEY: in 'Phase Transformations in Ferrous Alloys', (Eds. A. R. Marder and J. I. Goldstein), 201-236; 1984, Warrendale, TMS-AIME.
3. R. F. MEHL, C. S. BARRETT, D.W. SMITH: *Trans. AIME*, 1933, **105**, 215-258.
4. C. ZENER: *Trans. AIME*, 1946, **167**, 550-595.
5. J. W. CHRISTIAN: 'The Theory of Transformations in Metals and Alloys. Part 1', 19; 1975, Oxford, Pergamon Press.
6. M. HILLERT: *Jernkont. Ann.*, 1957, **141**, 757-764.
7. M. HILLERT: 'The Mechanism of Phase Transformation in Crystalline Solids, 231; 1969, London, Institute of Metals.
8. **K. HASHIGUCHI and J. S. KIRKALDY: Scandinavian Journal of Metallurgy, 1984, 13, 240-248.**
9. C. I. GARCÍA and A. J. DEARDO: *Metall. Trans.*, 1981, **12A**, 521-530.
10. G. R. SPEICH, V. A. DEMAREST and R. L. MILLER: *Metall. Trans.*, 1981, **12A**, 1419-1428.
11. M. M. SOUZA, J. R. C. GUIMARAES and K. K. CHAWLA: *Metall. Trans.*, 1982, **13A**, 575-579.
12. XUE-LING CAI, A. J. GARRAT-REED and W. S. OWEN: *Metall. Trans.*, 1985, **16A**, 543-557.
13. C. GARCÍA DE ANDRÉS, F. G. CABALLERO and C. CAPDEVILA: *Scripta Mater.*, 1998, **38**, 1835-1842.
14. G. R. SPEICH and A. SZIRMAE: *Trans. TMS-AIME*, 1969, **245**, 1063-1074.

15. A. ROOSZ, Z. GACSI and E. G. FUCHS: *Acta Metall.*, 1983, **31**, 509-517.
16. R. R. JUDD and H. W. PAXTON: *Trans. TMS-AIME*, 1968, **242**, 206-215.
17. M. HILLERT, K. NILSSON and L. E. TORNDAHL: *JISI*, 1971, **209**, 49-66.
18. D. P. DATTA and A. M. GOKHALE: *Metall. Trans.*, 1981, **12A**, 443-450.
19. E. NAVARA and R. HARRYSSON: *Scripta Metall.*, 1984, **18**, 605-610.
20. S. F. DIRNFELD, B. M. KOREVAAR and F. VAN'T SPIJKER: *Metall. Trans.*, 1974, **5**, 1437-1444.
21. E. E. UNDERWOOD: 'Quantitative Stereology', 73-75; 1970, Reading, Addison-Wesley Publishing Co.
22. S. A. SALTYSKOV: 'Stereometric Metallography', 2nd ed., ADC267-701; 1958, Moscow, Metallurgizdat.
23. R. T. DE HOFF and F. H. RHINES: 'Quantitative Stereology', 93; 1968, New York, McGraw-Hill.
24. F. S. LE PERA: *J. Metals*, 1980, **32**, 38-39.
25. M. P. PULS and J. S. KIRKALDY: *Metall. Trans.*, 1972, **3**, 2777-2796.
26. J. J. KRAMER, G. M. POUND, and R. F. MEHL: *Acta Metall.*, 1958, **6**, 763-771.
27. M. AVRAMI: *J. Chem. Phys.*, 1940, **8**, 212.
28. G. A. ROBERTS and R. F. MEHL: *Trans. ASM.*, 1943, **31**, 613-650.
29. R. G. KAMAT, E. B. HAWBOLT, L. C. BROWN and J. K. BRIMACOMBE: *Metall. Trans.*, 1992, **23A**, 2469-2482.
30. C. GARCÍA DE ANDRÉS, F. G. CABALLERO, C. CAPDEVILA and H. K. D. H. BHADOSHIA: *Scripta Mater.*, 1998, **39**, 791-796.
31. K. W. ANDREWS: *JISI*, 1965, **203**, 721-727.

Table 1 Isothermal Conditions Employed for the Formation of Fully Pearlitic Microstructures

Specimen	Temperature / K	Time / min
MORF1	948	45
MORF2	923	10
MORF3	798	60

Table 2 Morphological Characterization of Pearlite

Specimen	$\sigma_o / \mu\text{m}$	$S_v^{PP} \times 10^{-3} / \mu\text{m}^{-1}$	$a^P / \mu\text{m}$
MORF1	0.20±0.03	581±86	4.16±0.70
MORF2	0.08±0.01	1399±273	1.76±0.34
MORF3	0.06±0.01	1432±60	1.65±0.07

Table 3 Temperatures in K of Heating Interruption by Quenching

Morphology	T_a	T_b	T_c	T_d	T_e	T_f	T_g
MORF1	1014	1019	1023	1026	1041	1050	1060
MORF2	1002	1007	1012	1019	1026	1039	1049
MORF3	1001	1006	1008	1011	1019	1031	1041

a MORF1 (SEM), $\sigma_o=0.20 \mu\text{m}$; *b* MORF2 (SEM), $\sigma_o=0.08 \mu\text{m}$; *c* MORF3 (TEM), $\sigma_o=0.06 \mu\text{m}$

1 Electron micrographs of the three different morphologies of pearlite considered in this study

2 Temperatures selected from heating dilatometric curves to investigate the progress of reaustenitization from pearlite

3 Experimental values and theoretical variation of interlamellar spacing as a function of undercooling in a eutectoid steel

4 Experimental and calculated kinetics results for the formation of austenite inside pearlite under continuous heating conditions in eutectoid steel

a T_S , MORF1; *b* T_I , MORF1; *c* T_F , MORF1; *d* T_S , MORF2; *e* T_I , MORF2; *f* T_F , MORF2; *g* T_S , MORF3; *h* T_I , MORF3; *i* T_F , MORF3

5 Optical micrographs corresponding to the start (T_S), intermediate (T_I) and finish (T_F) temperatures of the re-austenitization process

a start temperature, T_S ; *b* finish temperature, T_F

6 Comparison of experimental and calculated critical temperatures of pearlite-to-austenite transformation in a eutectoid steel.