AUSTENITE TO FERRITE TRANSFORMATION KINETICS

DURING CONTINUOUS COOLING

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Abstract

The austenite decomposition has been investigated in a hypo-eutectoid plain carbon steel under continuous cooling conditions using a dilatometer and a Gleeble 1500 thermomechanical simulator. The experimental results were used to verify model calculations based on a fundamental approach for the dilute ternary system, Fe-C-Mn.

The austenite to ferrite transformation start temperature can be predicted from a nucleation model for slow cooling rates. The formation of ferrite nuclei takes place with equilibrium composition on austenite grain boundaries. The nuclei are assumed to have a pill box shape in accordance with minimal interfacial energy. For higher cooling rates, early growth has to be taken into account to describe the transformation start.

In contrast to nucleation, growth of the ferrite is characterized by paraequilibrium; i.e. only carbon can redistribute, whereas the diffusion of Mn is too slow to allow full equilibrium in the ternary system. However, Mn segregation to the moving ferrite-austenite interface has to be considered. The latter, in turn, exerts a solute drag effect on the boundary movement. Thus, growth kinetics is controlled by carbon diffusion in austenite modified by interfacial segregation of Mn. Employing a phenomenological segregation model, good agreement has been achieved with the measurements.

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Introduction

Phase transformations in steels have been investigated more extensively than those in any other material and many papers have been published since the 1930s. However, the characteristics of the entire transformation process are not yet fully understood. The factors affecting austenite decomposition are chemistry, initial austenite grain size, cooling rate and retained strain.

The present paper deals with the austenite-to-ferrite transformation in a low carbon, plain carbon steel under cooling conditions similar to those obtained on the run-out table of a hot strip mill. A phenomenological model is proposed to describe nucleation and growth of this type of phase transformation.

Experimental

To determine the kinetics of decomposition of austenite-to-ferrite and pearlite, dilatometric measurements have been performed on tubular specimens (diameter 8mm, wall thickness 1mm) of A36 plain carbon steel (0.17wt%C and 0.74wt%Mn) using the GLEEBLE 1500 thermomechanical simulator. The samples were austenitized for two minutes at 950°C resulting in a mean volumetric austenite grain size, d_γ, of 18μm; this grain size is representative of that attained after finish rolling under mill conditions. After this standardized pre-heating schedule, the transformation test was performed by applying the appropriate cooling regime. Three types of cooling procedures were employed with the cooling rate, φ, in austenite being measured at 850°C: controlled cooling for low φ (<10°C/s), air cooling (φ ≈20 °C/s) and He cooling for higher φ, where the He flow rate was varied to obtain a range of cooling rates extending from 30 to 250°C/s. In addition to the dilatometric dilation vs. time test results, the resulting microstructure was also quantified.

The experimental results can be summarized as follows. For all cooling rates, a ferritic-pearlitic microstructure was formed with a ferrite fraction, F, of approximately 0.8. Up to a cooling rate of 80°C/s, polygonal ferrite formed having a grain size, d_α (in μm), which decreased with increasing φ,

$$d_\alpha = 12.33\varphi^{-0.14} \tag{1}$$

Further increase of cooling rate encouraged the formation of Widmanstätten ferrite, making any ferrite grain size measurement uninterpretable. In all cases, ferrite nucleated at the austenite boundaries. To characterize the transformation start and finish, the temperatures $T_{0.05}$ (5% transformed) and $T_{0.95}$ (95% transformed) have been determined from the dilatometric response. As shown in Figure 1, both temperatures decreased with increasing cooling rate.

Nucleation

The nucleation model of Enomoto and Aaronson [1] has been extended to continuous cooling conditions. The classical nucleation rate, J, can be expressed as

$$J = \eta \frac{2D\Omega x}{a^4(3kT)^{1/2}} \exp\left(-\frac{\varsigma}{\Delta G^2 kT}\right) \tag{2}$$

where ΔG is the driving force for ferrite nucleation, D is the relevant (i.e., bulk or grain boundary) diffusivity of the rate determining species with an atomic fraction x, Ω is the atomic volume of ferrite, a the average lattice parameter of ferrite and austenite, and η and ς represent parameters which depend on the potential nucleation site density and interfacial energies. Enomoto and Aaronson investigated in detail ferrite nucleation at austenite grain

boundaries and favor a pillbox nucleus model over those based solely on spherical caps. Within the coherent pillbox model,

$$\eta = N(\gamma_c + \gamma_o - \gamma_{gb})^{1/2} \qquad (3)$$

and

$$\varsigma = 4\pi\gamma_e^2(\gamma_c + \gamma_o - \gamma_{gb}) \qquad (4)$$

is obtained, where the two parameters η and ς are expressed by N, the area density of the atomic sites available for nucleation, γ_{gb}, the grain boundary energy, and the interfacial energies γ_c, γ_e and γ_o of the various partially or fully coherent surfaces of the pillbox.

The nucleation temperature, T_N, obtained during continuous cooling can be determined from

$$\Theta = \int_{T_N}^{T_{Ae3}} \frac{J(T)}{\varphi(T)} dT \qquad (5)$$

where the temperature dependence of the cooling rate is considered. T_{Ae3} is the temperature below which single phase austenite becomes thermodynamically unstable; $T_{Ae3}=824°C$ for the A36 steel. The critical nucleation density, Θ, can be evaluated from the ferrite grain size by

$$\Theta = \frac{Fd_\gamma}{\pi d_\alpha^3} \qquad (6)$$

assuming that the number of ferrite grains per unit volume in the final microstructure equals the number of nuclei per unit volume at transformation start.

The model has been applied to predict T_N for the austenite-to-ferrite transformation in A36 steel using the thermodynamic data of Kirkaldy and Baganis [2] to quantify ΔG for the formation of ferrite nuclei with equilibrium composition. The experimental observations for slow cooling rate ($\varphi=1°C/s$), where the transformation start and the nucleation temperature are practically identical, indicate T_N to be approximately 760°C, which is supported by isothermal tests. Assuming grain boundary diffusion of Mn, $D=0.89cm^2s^{-1}exp(-1.66eV/kT)$ [4], to be rate controlling, the model predicts $T_N = 757°C$ at a cooling rate $\varphi = 1°C/s$ for the parameters $\eta =10^{10}$ and $\varsigma=8.8\times10^{-5}$ $(Jm^{-2})^3$. Since the difference of interfacial and grain boundary energies, $\gamma_c+\gamma_o-\gamma_{gb}$, can be assumed to be in the range of 0.005-0.008Jm^{-2} [1], reasonable values of N (approximately $10^{11}m^{-2}$) and γ_e (0.030-0.037Jm^{-2}) are obtained. The energy of the coherent α-γ interface is in good agreement with the literature, where 0.018Jm^{-2} [3] and 0.05-0.09Jm^{-2} [1] have been reported for γ_e in Fe-0.13wt%C and Fe-0.11wt%C-3.1wt%Mn, respectively, confirming the trend to larger γ_e with increasing Mn content.

Since η and ς are independent of cooling rate, the above model can also be used to predict T_N for higher cooling rates. The predicted T_N values are shown by the dashed line in Figure 1. Comparison with the experimental transformation start temperature shows that T_N does not decrease as rapidly with cooling rate as does $T_{0.05}$. Since some growth is necessary to detect the transformation start, the time required for early growth accounts for the increasing difference between T_N and $T_{0.05}$, as shown below.

Ferrite Growth

Assuming nucleation site saturation after 5% ferrite has formed, a condition supported by metallographic observations, the transformation kinetics is solely characterized by ferrite growth until the final ferrite fraction is attained and pearlite forms. Carbon diffusion in austenite is generally thought to control the rate of ferrite growth. Kamat et al. [5] describe the diffusion process assuming α-γ interfacial equilibrium, spherical geometry, where the diameter is equal to the initial austenite grain size, and no flux conditions at the center of the spherical grain. This model gives reasonable agreement with isothermal transformation kinetics in low manganese steels. Extending this model to continuous cooling conditions, a much more rapid growth rate is predicted in the A36 steel than is experimentally observed. This agrees with earlier findings that Mn slows the transformation rate down, presumably by solute drag [6,7]. Thus, the present approach incorporates the solute drag effect of Mn into the carbon diffusion model. On basis of the experimental results, the growth process can be subdivided into three stages.

The first stage, i.e. early growth, is characterized by a noticeable delay, as seen from the differences between T_N and $T_{0.05}$. Nucleation of ferrite requires the redistribution of both, C and Mn. Since the equilibrium Mn concentration in ferrite is lower than that in austenite and Mn diffusion in austenite is a comparatively slow process, enrichment of Mn can initially be expected at the interfaces of the ferrite nuclei. Mn atoms, originally segregated to the prior austenite grain boundary area at which nucleation is occurring, further enhance the initial Mn enrichment at the α-γ interface. Assuming local equilibrium, the higher interfacial Mn concentration reduces the interfacial C concentration, which may result in a decreasing gradient for C diffusion. Thus, initially some Mn diffusion is required to unpin the interface by allowing C diffusion to take place.

The desegregation of Mn can be described by one-dimensional diffusion from its enrichment, $s(T_N)$, deposited at T_N in the interface plane [8]

$$s(T) = s(T_N) \frac{\int_0^b \exp\{-x^2 / 4\Gamma(T)\}dx}{(\pi\Gamma(T))^{1/2}} \qquad (7)$$

where $s(T)$ is the segregation factor of Mn, b is the magnitude of the Burgers vector representing the width of the semi-coherent α-γ interface and Γ is related to the Mn diffusion coefficient in austenite, D_{Mn}[9], by

$$\Gamma(T) = \int_{T(t)}^{T_N} \frac{D_{Mn}(T')dT'}{\varphi(T')} \qquad (8)$$

The transformation start temperature, T_S, is estimated by assuming that the segregation factor $s(T)$ must decrease to a level which allows the formation of a carbon concentration gradient in front of the interface, i.e., the local equilibrium interfacial carbon concentration has to be bigger than the average carbon bulk concentration. Applying this criterion, an initial segregation factor of approximately 25 has been estimated in the present case and the respective values for T_S are shown by the dotted line in Figure 1.

In the second stage of the growth process, which begins at T_S, C diffusion modified by Mn solute drag is rate controlling, with the carbon diffusivity given by Agren [10]. The solute drag effect is included by selecting the appropriate local equilibrium condition at the α-γ

interface, accounting for the higher interfacial Mn concentration. A steady-state Mn enrichment is retained at the moving interface. The steady state segregation factor is given by

$$s_s = \exp(E(T)/kT) \qquad (9)$$

where $E(T)$ is the effective segregation energy; this temperature dependent energy accounts for an increasing amount of Mn being able to follow the interface movement with increasing temperature. As shown in Figure 2, this modification of the carbon diffusion model leads to an adequate description of the observed ferrite growth when the segregation energy (in eV)

$$E(T) = 0.25 - 0.0012(T_{Ae3} - T) \qquad (10)$$

is taken.

Obviously, solute drag conditions obtained in the later stages of the transformation are not included in this model. With the onset of soft impingement, the rate of interface movement decreases, enabling more Mn to desegregate and thereby reducing its solute drag effect. Later on, the interface velocity is low enough and the redistribution of C in the austenite is sufficient for the formation of pearlite [11]. Further studies are under way to quantify these critical velocities. The present results indicate that for the low cooling rate (1° C/s), where the transformation takes place at comparatively high temperatures, the reduction of the solute drag force in the later stages of ferrite growth is significant. For higher cooling rates, which are typical of the run-out table in the finishing mill, the solute drag is rather small because of the high undercooling and a more accurate description of the reduced solute drag in the final stage of ferrite growth is less important.

Conclusions

The austenite-to-ferrite transformation in a low carbon steel is described with a phenomenological nucleation and growth model for industrial cooling conditions. The nucleation temperature is estimated from the nucleation model of Enomoto and Aaronson. Initial growth is very slow since Mn diffusion is required to reduce the Mn enrichment at the α-γ interface, which results from the nucleation process and prior Mn grain boundary segregation. With Mn desegregation the growth rate quickly increases to a level limited by C diffusion in austenite and Mn solute drag at the moving α-γ interface, an effect which increases with temperature.

These findings agree with earlier investigations that the austenite decomposition can be inhibited by adding Mn, with the delay being more pronounced at higher temperatures. Further studies are in progress to validate the proposed model for different steel chemistries.

Acknowledgments

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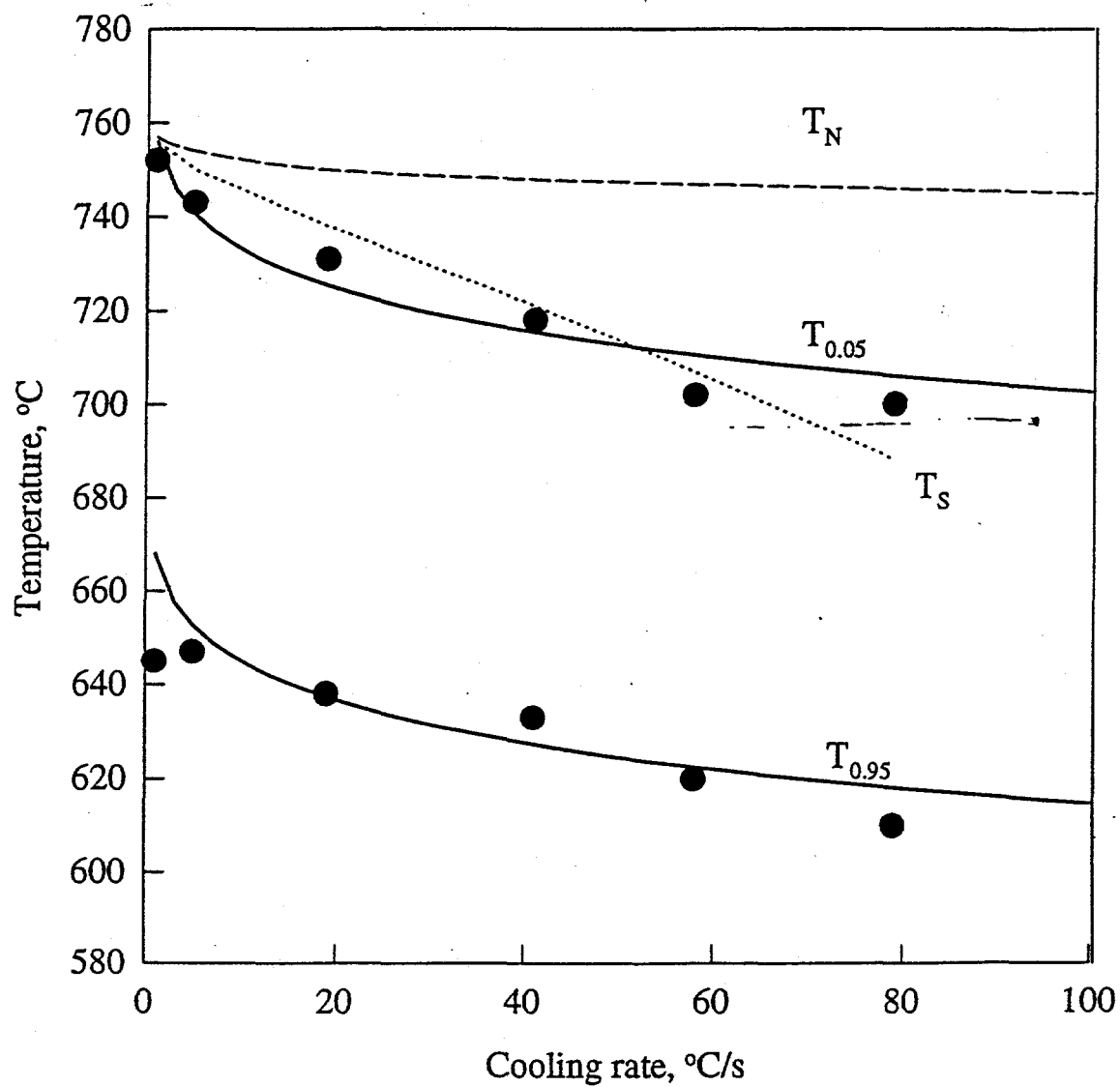11. M. Umemoto et al., ISIJ International, 32 (1992) 306-315.

Figure 1

Experimental transformation start ($T_{0.05}$) and finish ($T_{0.95}$) and predicted nucleation (T_N) and start (T_S) temperatures.
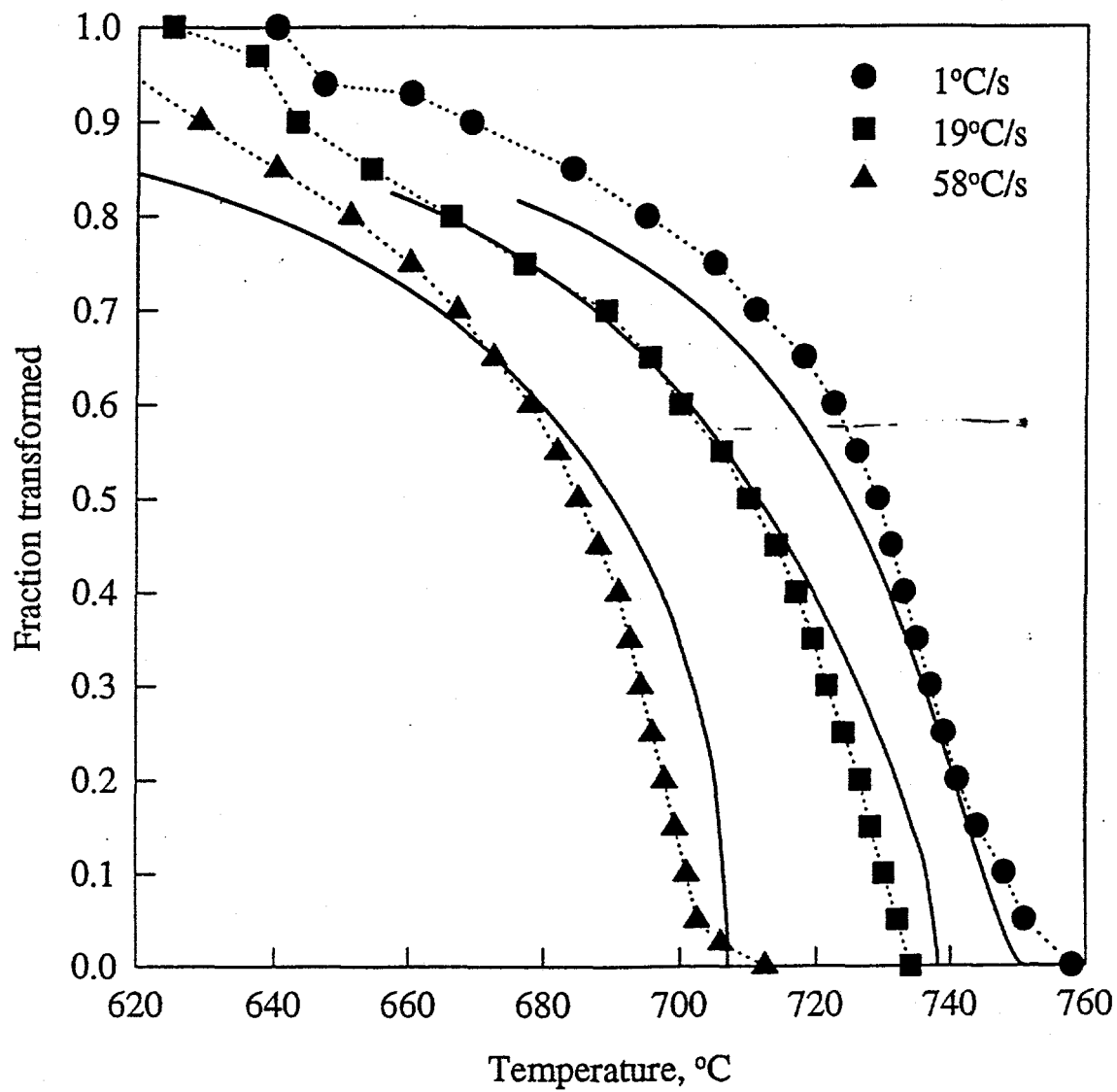
Figure 2
Comparison of model predictions for ferrite growth (solid lines) with the transformation kinetics in the A36 steel.