FERRITE FORMATION IN HETEROGENEOUS DUAL-PHASE STEELS

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Introduction

Many properties of dual-phase steels depend on the volume fractions V and $(1-V)$ of ferrite (α) and martensite, respectively, in the microstructure. The desired microstructure can be obtained by the isothermal transformation of austenite (γ) in the $\alpha+\gamma$ phase field, followed by rapid cooling to ambient temperature so that any remaining γ decomposes martensitically. This work is an investigation of the paraequilibrium (1) formation of ferrite in "banded" Fe-Mn-C steels – i.e., steels which are compositionally heterogeneous, as is usually the case in commercial practice (henceforth, all V refer to α formed by the paraequilibrium mechanism). For heterogeneous steels, it is not obvious how the maximum volume fraction V_{ht} of α that can be obtained by isothermal transformation (at a temperature T) in the $\alpha+\gamma$ phase field should differ from V_{hm}, the corresponding maximum volume fraction of ferrite obtainable in a homogeneous steel of the same mean composition. For homogeneous steels, the lever rule suggests that V_{hm} should vary linearly with $\bar{x}$, the average carbon content of the steel concerned, when isothermally transformed at any particular T and there is some experimental evidence (2) to show that this relation does not hold for banded steels.

Theory, Results and Discussion

The analysis is based on Fe-Mn-C steels containing a one dimensional, constant gradient of Mn, a gradient which is almost unaffected by the paraequilibrium growth of α. This is illustrated in Fig.1 (curve A), with the co-ordinate Z defined to be normal to the planes of constant composition (i.e., the planes normal to the plane of the diagram) in the specimen. Z is normalised to have a maximum value of unity, equivalent to the total length of the specimen in the Z direction. Ferrite formation should first occur in the low Mn region (Z=0) and on average, the transformation can be visualised to take place by the motion of a planar γ/α interface whose normal is in the Z direction. The position of this interface, with repect to Z, at any instant defines the volume fraction V of ferrite since $V \equiv Z$.

In reality, the variation of Mn in banded steels is much more complex than the linear variation indicated in Fig.1. It might be considered more appropriate to represent the banding in terms of a composition wave along Z. If such a wave is taken to be a triangular wave of wavelength λ (normalised with respect to the specimen length in the Z direction), then $\lambda=2$ for curve A and $\lambda=1$ for curve B of Fig.1. For case B, transformation should begin at Z=0 and at Z=1, the interfaces moving in the +Z and -Z directions, respectively. If the reaction ceases when the interface encounters a Mn level of $Mn=Mn_1$, then the volume fraction of α obtained for case A is z_2 and for case B is $z_1 + (1 - z_3)=z_2$.

FIG.1

Simulation of Banding. The co-ordinate Z is defined to be normal to the planes of constant composition and is normalised with respect to the specimen length in the Z direction. The curves A and B represent Mn concentration gradients in the banded steel.
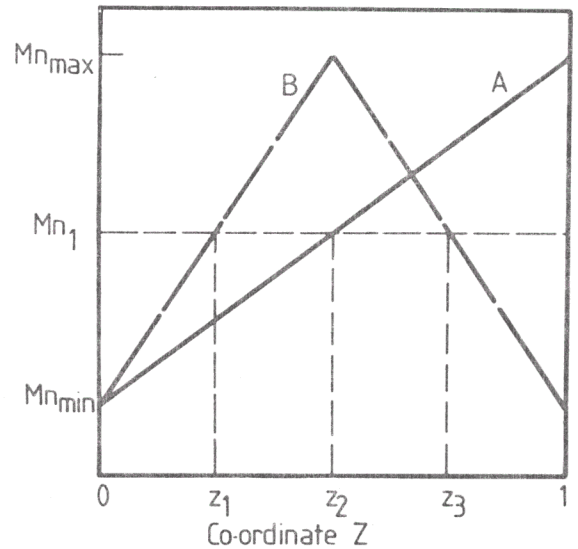
It follows that the final volume fraction of α obtained at any T is not a function of λ and this conclusion holds even if the composition wave is a sine wave (if λ becomes very small, as in spinodal decomposition, then other complications arise - see Ref.3). The remaining analysis is carried out for case A.
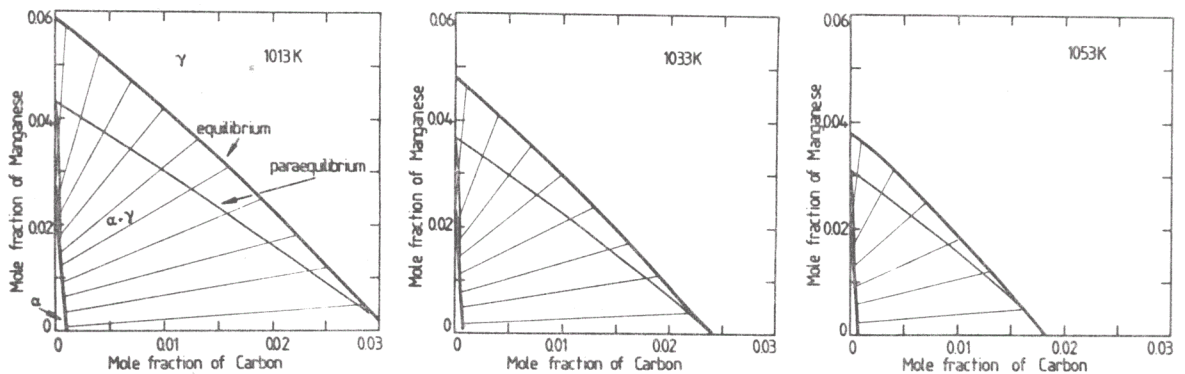

FIG.2

Calculated isothermal sections of the Fe-Mn-C phase diagram, including the equilibrium and paraequilibrium phase boundaries. The tie-lines are for the equilibrium case, since those for paraequilibrium diagram are all nearly parallel to the horizontal axis.

If the small difference in the densities of ferrite and austenite is ignored, the carbon content of the residual austenite at any stage of transformation (assuming a uniform distribution of carbon) is given by

$$x_\gamma = (\bar{x} - Vx_\alpha^{\alpha\gamma})/(1 - V) \qquad \qquad .....(1)$$

where $x_\alpha^{\alpha\gamma}$ is the paraequilibrium carbon content of the α at the α/γ interface, a function of the Mn level and of T. The corresponding carbon content of the γ at the α/γ interface is $x_\gamma^{\gamma\alpha}$. Both these concentrations can be obtained from the paraequilibrium phase boundaries given in Fig.2, which is

calculated as in Ref.4. The formation of ferrite is expected to cease when $x_\gamma = x_\gamma^{\gamma\alpha}$. For the heterogeneous alloy under consideration, $x_\gamma^{\gamma\alpha}$ varies with the position of the interface, since the Mn level also varies with Z. $x_\gamma^{\gamma\alpha}$ is therefore a function of V and the condition for reaction to cease may be written as

$$x_\gamma^{\gamma\alpha}\{V\} = x_\gamma\{V\} \qquad\qquad \dots(2)$$

where $x_\gamma^{\gamma\alpha}$ and x_γ are implied to be functions of V, by using the curly brackets. The two sides of this equation can be separately evaluated (using Fig.2 and eq.1) and each plotted as a function of V, as shown in Fig.3, where the calculations are carried out for steels with three different carbon concentrations, and for isothermal transformation at three different temperatures in the $\alpha+\gamma$ phase field. Curves such as 'ab' represent $x_\gamma^{\gamma\alpha}\{V\}$ and the curves 'cd' correspond to $x_\gamma\{V\}$. Each intersection is a solution to eq.2, and gives a value of V_{ht}. The Mn concentration at the position where the interface stops is $[0.8+(1.2-0.8)V_{ht}]$ since Mn_{min} and Mn_{max} (see Fig.1) were taken to be 0.8 and 1.2 wt.pct., respectively. Each open circle in Fig.3 represents the value of V_{hm} for a homogeneous steel containing 1.5 wt.pct. Mn which is equal to the mean Mn level, $\overline{Mn}$, of the heterogeneous steel. To note the difference between the transformation behaviour in homogeneous and heterogeneous steels, each open circle should be compared with the nearest intersection between curves of type ab and cd – this gives corresponding values of V_{hm} and V_{ht}. The following points can be noted from the results of Fig.3:

1) It is evident that there are circumstances when $V_{ht} > V_{hm}$, and also cases where $V_{ht} < V_{hm}$. The former occurs when the interface stops at a point where the Mn level is less than $\overline{Mn}$ and the latter when this Mn level is greater than $\overline{Mn}$. When heterogeneous steels are heat-treated at relatively large undercoolings, such that the volume fraction of ferrite obtained is high enough to allow the tranformation interfaces to stop at points where $Mn>\overline{Mn}$, the actual volume fraction of α obtained will be less than that expected for a homogeneous alloy. There will be no difference in V_{ht} and V_{hm} when the Mn concentration at the interface equals $\overline{Mn}$.

2) When $V_{ht} > V_{hm}$, the carbon content of the residual austenite will be higher than that expected for a homogeneous steel, and vice-versa. This might significantly influence the mechanical properties of the dual-phase steel concerned.

3) The linear variation of $(1 - V_{hm})$ with $\overline{x}$ implied by the lever rule will clearly not be valid for heterogeneous steels; $(V_{ht} - V_{hm})$ depends on the absolute value of V_{ht}, which is in turn a function of $\overline{x}$. For any particular T, higher carbon steels lead to the formation of lower quantities of α before reaction ceases, and the interface is then more likely to stop at a position where $Mn<\overline{Mn}$. This would lead to a higher than expected quantity of α so that the curve $(1-V_{ht})$ versus $\overline{x}$ would tend to have a decreasing positive slope with increasing $\overline{x}$, in agreement with the work of Speich and Miller.

Acknowledgement

References

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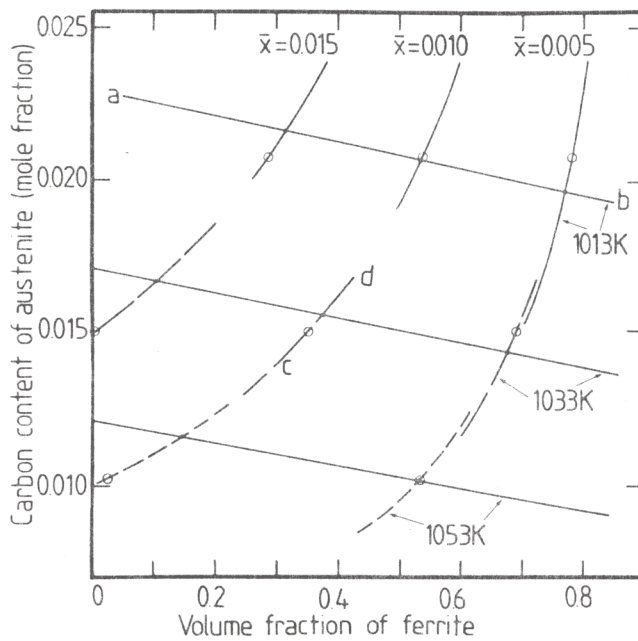4) J.B. Gilmour, G.R. Purdy and J.S. Kirkaldy, Metall. Trans., 3, 1455 (1972)

FIG.3

Plot to determine the solutions of equation 2.