Carbon content of retained austenite in quenched steels

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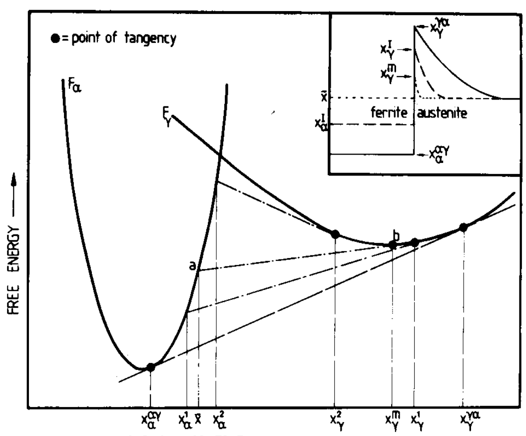
Much effort has recently been directed towards the determination of the carbon content of retained austenite, i.e. x_γ, in directly quenched,[1] or quenched and tempered[2] low alloy steels. The results indicate that x_γ may be significantly higher than the average carbon content $\bar{x}$ of the alloy concerned. The interpretation of such data is difficult, but the present paper is concerned with the possibility that some redistribution of carbon may occur *during* the transformation of austenite to what is generally referred to as 'lath martensite' in low alloy steels; the transformation may not be truly diffusionless and could perhaps involve the displacive growth of only partially supersaturated laths of martensite. The carbon content of such laths during growth, i.e. x_α^I, may be less than $\bar{x}$, although greater than the paraequilibrium level $x_\alpha^{\alpha\gamma}$ given by the common tangent construction of Fig. 1, the remainder being pushed ahead of the transformation interface, giving rise to a diffusion controlled growth rate under steady state conditions.

Under these circumstances the carbon contents at the interfaces at any stage of steady state growth, in each of the phases, would be thermodynamically linked and specified by means of a tie line (such as ab, which links the general set of compositions labelled x_α^I and x_γ^I, in Fig. 1). In Fig. 1 the tie line $x_\alpha^2 - x_\gamma^2$ illustrates why x_γ^1 cannot be less than x_γ^m for an alloy with average carbon content $\bar{x}$, since x_α^1 cannot exceed $\bar{x}$. The inset in Fig. 1 illustrates the types of composition profile to be expected across the transformation interface, under conditions of steady state growth, with carbon content and position relative to the interface plotted on the ordinate and abscissa respectively.

It is necessary to comment on the applicability of a tie line construction to the case of non-equilibrium segregation being discussed here. Suppose we consider (Fig. 1) the diffusion controlled growth of ferrite of composition x_α^1. It is then thermodynamically possible for any austenite whose composition falls within the range x_γ^1 to $\bar{x}$ to decompose into ferrite of composition x_α^1. However, because the analysis is restricted to diffusion controlled growth, the composition falls within the range x_γ^1–$\bar{x}$ to decompose itself (at steady state) to the maximum allowable carbon content consistent with the presence of x_α^1 amount of carbon in the ferrite. This maximum composition is of course x_γ^1 justifying the use of the tie line $x_\alpha^1 - x_\gamma^1$ of Fig. 1.

If the corresponding tie line compositions are written as x_α^I and x_γ^I, the reaction should cease when $x_\gamma = x_\gamma^I$. However, this may not be true if for some reason, such as mechanical stabilization, the reaction should cease before the matrix reached a uniform composition x_γ^I. For the present we will ignore this important difficulty, but as will be seen later, it introduces uncertainty into one of the conclusions that follow from the analysis to be presented. Bhadeshia and Waugh[3] have demonstrated that for growth involving a partial supersaturation, the minimum possible value of x_γ^I, and hence x_γ at reaction termination, equals x_γ^m (*see* Fig. 1), since this corresponds to the austenite tie line composition when the carbon concentration of the martensite, during growth, is only infinitesimally less than $\bar{x}$. This follows from the fact that the carbon content of the martensite cannot exceed $\bar{x}$ under any circumstances.[3] Hence, any experimentally observed value of x_γ (at reaction termination) which falls below x_γ^m indicates inconsistency with the hypothesis that the redistribution of carbon, between the lath martensite and residual austenite occurred during the growth of the former; such an observation therefore implies that the carbon may have diffused into the remaining austenite after the actual transformation event corresponding to the formation of true martensite.

The application of the above theory to the experiments of Refs. 1 and 2 is complicated, since the martensite involved was actually formed over a range of transformation



1 Schematic free energy curves (taken from Ref. 3) for ferrite and austenite with same substitutional alloying element content, indicating definition of tie lines involved in growth of ferrite with only partial supersaturation of carbon: $x_\alpha^{\alpha\gamma}$ and $x_\gamma^{\gamma\alpha}$ are the paraequilibrium carbon concentrations in ferrite and in austenite respectively (*see* text for details)

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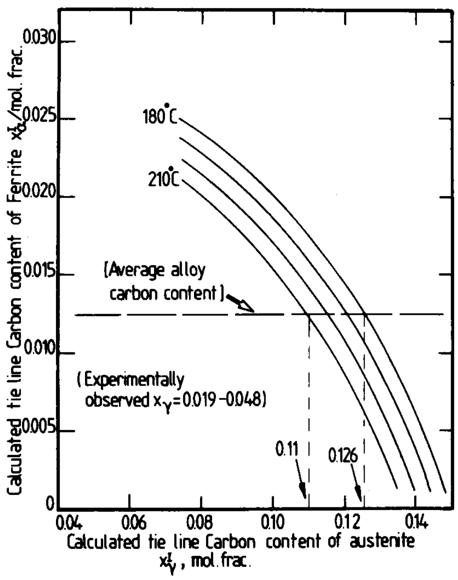
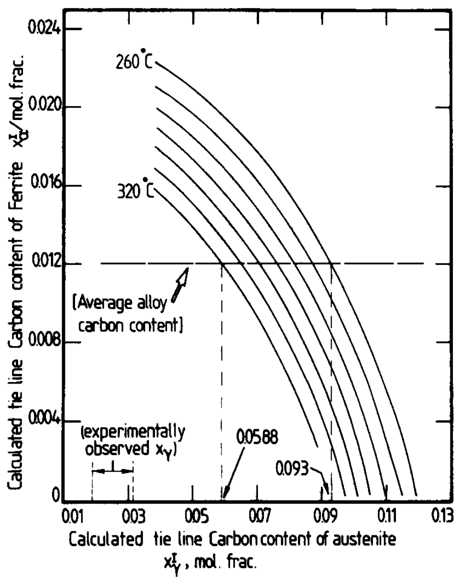


2 **Graph of x_α^I v. x_γ^I: experimental x_γ data,[1] determined by lattice imaging technique, refers to retained austenite in directly quenched Fe–4·0Cr–5·0Ni–0·27C (wt-%) steel**



3 **Graph of x_α^I v. x_γ^I: experimental x_γ data,[2] determined using atom probe, refers to retained austenite in Fe–0·26C–2·99Cr–1·98Mn–0·50Mo–0·07Si (wt-%) steel quenched and subsequently tempered at 200°C for 2 h**

temperatures between M_s and 'M_f'. In addition, one of the steels concerned[2] had been given a low temperature (2 h at 200°C) tempering treatment. It is believed that this tempering treatment can only lead to an increase in x_γ, so that the decision on the transformation mechanism, as deduced from the above theory, can only become ambiguous if x_γ happened to exceed x_γ^m. In addition, the significance of this tempering treatment is probably limited as far as the present analysis is concerned, since the steel concerned has an M_f of 260°C, well above the 200°C imposed temper. Indeed, the role of autotempering in causing compositional variations cannot be properly established for either of the steels concerned. We emphasize, however, that these difficulties of interpretation only become significant if $x_\gamma > x_\gamma^m$.

The fact that the martensite formed over a range of temperatures can be taken into account by computing x_γ^m for the range M_s–M_f. Figures 2 and 3 show a series of such calculations with each curve, representing calculations at a specified temperature, being a plot of x_α^I *v.* x_γ^I, for various levels of imposed x_α^I. The x_γ^m values corresponding to each curve can be read off as the value of x_γ^I at $x_\alpha^I = \bar{x}$. The details for the calculations are given in Ref. 3, but the stored energy of martensite has been taken to be 700 J mol^{-1}, according to Ref. 4. It is clear from Figs. 2 and 3 that the experimentally observed values of x_γ (from Refs. 1 and 2) fall well below the calculated x_γ^m values. The results would indicate that the carbon enrichment of the austenite occurred after the actual formation of the martensite, the growth of the latter phase being diffusionless. However, as pointed out above, this conclusion can only be considered to be established if it is justified to assume that diffusion controlled (displacive) growth would continue until the matrix reached a uniform composition x_γ^I. Despite this uncertainty, the analysis definitely shows that the experimental results on martensite may only be claimed to be consistent (without proving) with carbon diffusion *during* growth if, and only if, some factor other than carbon redistribution, e.g. mechanical stabilization, prevents the residual austenite from undergoing displacive transformation.

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