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Reaustenitisation from bainite

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SYNOPSIS

Reaustenitisation by the diffusional growth of austenite from a microstructure of bainitic ferrite (α_b) and austenite, has been investigated and the results are explained theoretically.

INTRODUCTION

This work is limited to a study of *isothermal* reaustenitisation on a chemically homogeneous alloy in which the starting microstructure already contains austenite, whose nucleation is consequently unnecessary. The work is part of a project on welding and the alloy used is a homogenised weld deposit of composition Fe-0.06C-0.27Si-1.84Mn-2.48Ni-0.20Mo-0.0032O-0.01Al-0.02Ti, wt.%. The oxygen is tied up in the form of oxide inclusions. Alloy preparation and other experimental details have been presented elsewhere [1]; the dilatometric data were collected using a computer, at 0.01s intervals since transformation rates can be rather high.

RESULTS AND DISCUSSION

Many investigations have indicated that α_b plates grow by a diffusionless and displacive transformation mechanism but immediately afterwards, the excess carbon is partitioned into the residual austenite. The transformation exhibits an *incomplete reaction phenomenon* in which reaction ceases well before the residual austenite achieves its equilibrium carbon concentration. It stops when the carbon concentration of the residual austenite (x_γ) reaches the T_o' curve on the phase diagram. The T_o curve defines temperatures at which α and γ of the same composition have identical free energy [2-6]. The T_o' curve allows additionally for the fact that the stored energy of α_b is $\approx$ 400J/mol [3,5,7]. Thus, composition invariant transformation is only possible if x_γ is less than $x_{T_o'}$, the carbon concentration specified by the T_o' curve for the isothermal transformation temperature concerned.

Our aim was to study isothermal reaustenitisation, beginning with a microstructure of bainitic ferrite and austenite. Specimens austenitised at 950°C for 10 min (to give a small γ grain size) were isothermally transformed to bainite (Fig. 1) at a temperature $T = 460^oC$ for 30 min (i.e., longer than needed to complete the α_b reaction [1]) and then, without cooling below 460°C, rapidly up-quenched to a temperature T_γ for isothermal reaustenitisation. It has been demonstrated that for the alloy used, the bainite transformation ceases when x_γ reaches the T_o' curve, whether the bainite is nucleated at γ grain boundaries or intragranularly nucleated at inclusions [1]. The precipitation of cementite (θ) from γ is very slow for this alloy, when compared with the time required for the α_b transformation to cease. For example, no θ is found long after the completion of bainite transformation at 460°C and even when this mixture of bainitic ferrite and carbon-enriched residual γ is annealed by raising the temperature to 600°C (directly from 460°C), θ only precipitates from γ after $\approx$ 2 hours (Fig. 2). Hence, cementite precipitation does not complicate the interpretation of the results, reaustenitisation being from a mixture of just bainitic ferrite and carbon-enriched austenite.

For all the results reported here, the dilatometer output showed that there was no decomposition of austenite during heating. In these circumstances, reaustenitisation involves just the growth of γ and nucleation can be ignored. This has been confirmed by detailed electron microscopy to be reported elsewhere.

ISOTHERMAL REAUSTENITISATION

The first detectable growth of austenite was found to occur at $T_\gamma = 680^oC$ (Fig. 3). In all cases the transformation rate was initially rapid, but decreased with time so that the specimen length eventually stopped changing, as transformation ceased. Fig. 4 shows that ΔL_m (the maximum relative length change obtained at any T_γ) increases as T_γ changes from 680°C to 760°C and then stays essentially constant with further increase in T_γ.

A small amount of austenite growth occurred during the up-quench, when attempts were made to isothermally reaustenitise specimens at temperatures above 735°C. Its contribution to ΔL_m has been included in the data of Fig. 4; if the low-temperature part of the length versus temperature curve is extrapolated to T_γ, the vertical difference between the extrapolated curve and the actual curve gives the length change due to transformation during the up-quench, which should be added to any length change due to isothermal transformation at T_γ. The results show that below 760°C, reaustenitisation is incomplete, the alloy becoming fully austenitic only above this temperature. The maximum degree of transformation to austenite increases from nearly zero at 680°C to complete reverse transformation at 760°C and above. The rate of the reverse $\alpha \rightarrow \gamma$ transformation increases with T_γ (Fig. 3).

Table I: Microanalytical Data (wt.%, error $\approx \pm 0.2$)

t min,	T_γ °C	Mn γ	Mn α	Ni γ	Ni α
10	680	2.50	1.73	2.98	2.45
10	730	2.29	1.36	2.80	2.24
0.67	760	1.75	1.72	2.66	2.62
120	710	2.39	1.09	3.26	1.96
120	735	2.17	1.23	3.00	1.94
120	760	1.83	1.76	2.57	2.65

The microanalytical data (Table I) show that for low T_γ, substitutional alloying additions redistribute during the $\alpha \rightarrow \gamma$ transformation, even at low transformation times (t). For low t, the partitioning of alloying elements, as indicated by the deviation of the partition coefficient k_i from unity ($k_i = x_i^\gamma / x_i^\alpha$) increases with decreasing T_γ. This is consistent with the fact that at low-T_γ, the redistribution of substitutional alloying elements is a thermodynamic necessity [8,9]. As the driving force for reaustenitisation increases, the transformation tends

towards paraequilibrium or negligible-partitioning-local equilibrium (these concepts have been explained elsewhere [10]). This is illustrated by the data for 760°C, even for times as short as 40s. As t increases for a given T_γ, the partition coefficient k_i changes even though the volume fraction of austenite undergoes negligible change. Thus the concentrations of alloying elements at the interface during the growth of γ are not equilibrium concentrations.

The experiments emphasise that all compositions of γ which allow a reduction in free energy during growth can in principle grow from ferrite, not just those which lead to equilibrium at the interface. However, since the present work is concerned with the *maximum* degree of transformation, it is the equilibrium compositions which are of interest.

THEORETICAL ANALYSIS

Since after the diffusionless growth of bainite, carbon is rapidly and spontaneously redistributed into the residual austenite with an accompanying reduction in free energy, the $\alpha_b \rightarrow \gamma$ transformation in its original form is irreversible. Reaustenitisation is therefore different from the case of reverse transformation from martensite to austenite in for example, shape memory alloys.

The carbon concentration x'_γ of the austenite when the formation of bainite ceases at T_a, is given by (point marked **a** in Fig. 5):

$$x'_\gamma = x_{To}\{T_a\} \quad (1)$$

note that:

$$x'_\gamma << x_{Ae3}\{T_a\} \quad (2)$$

where $x_{Ae3}\{T_a\}$ is marked **b** in Fig. 5 (in all equations, the use of braces implies functional relations).

Thus, although the formation of *bainite* ceases at T_a, because $x'_\gamma << x_{Ae3}\{T_a\}$, the driving force for austenite to transform *diffusionally* to ferrite is still negative. This remains the case until the temperature T is high enough (i.e., $T=T_{\gamma 1}$) to satisfy the equation:

$$x'_\gamma = x_{Ae3}\{T_{\gamma 1}\} \quad (3)$$

Hence, reaustenitisation will first occur at a temperature $T_{\gamma 1}$, as indicated on Fig. 5 (marked **c**), and as observed experimentally. This is a consequence of the mechanism of the bainite transformation, which does not allow the transformation to reach completion. If this were not the case, then the lever rule demands that the temperature need only be raised infinitesimally above T_a in order for the reverse $\alpha \rightarrow \gamma$ transformation to be thermodynamically possible!

The theory predicts that for $T_\gamma > T_{\gamma 1}$, the $\alpha_b \rightarrow \gamma$ transformation should cease when x_γ (initially x'_γ) reaches the Ae3 curve, i.e., when

$$x_\gamma = x_{Ae3}\{T_\gamma\} \quad (4)$$

The equilibrium volume fraction of γ (at T_γ), is:

$$V_\gamma\{T_\gamma\} = \bar{x}/x_{Ae3}\{T_\gamma\} \quad (5)$$

assuming that the carbon concentration of α_b is negligible and that $x_{Ae3}\{T_\gamma\} > \bar{x}$. When $x_{Ae3}\{T_\gamma\} = \bar{x}$, the alloy eventually becomes fully austenitic (point **d**, Fig. 5), and if this condition is satisfied at $T_\gamma = T_{\gamma 2}$, then for all $T_\gamma > T_{\gamma 2}$, the alloy transforms completely to austenite.

These concepts explain the dilatometric data (Fig. 4) in which the degree of $\alpha \rightarrow \gamma$ transformation increases (from ≈ zero at 680°C) with the temperature of isothermal reaustenitisation, until the temperature $760°C = T_{\gamma 2}$ where the alloy transforms completely to austenite.

CONCLUSIONS

Isothermal reaustenitisation, beginning with a microstructure of just bainitic ferrite and austenite has been studied for circumstances where the nucleation of austenite is not necessary. Because the bainite transformation stops before the residual austenite achieves its equilibrium composition, reaustenitisation occurs only when the alloy is heated to a temperature where the volume fraction of austenite is less than its equilibrium volume fraction. This temperature may be below the Ae3 temperature of the alloy as a whole, but well in excess of the corresponding Ae1 temperature. Complete transformation to austenite only occurs when the alloy is heated to a temperture greater than the Ae3 temperature of the alloy. At all intermediate temperatures, the reverse $\alpha \rightarrow \gamma$ transformation terminates before the alloy becomes fully austenitic, with the volume fraction of austenite increasing with increasing T_γ.

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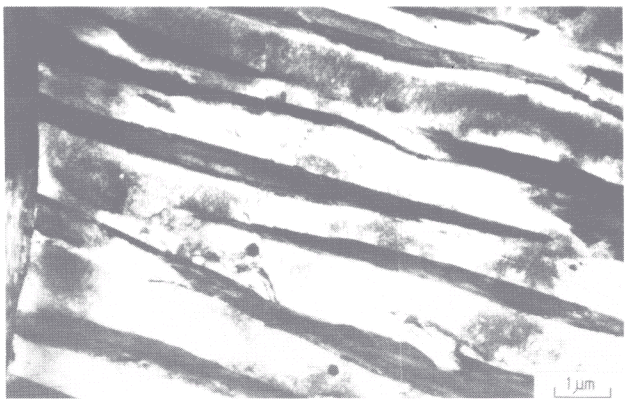
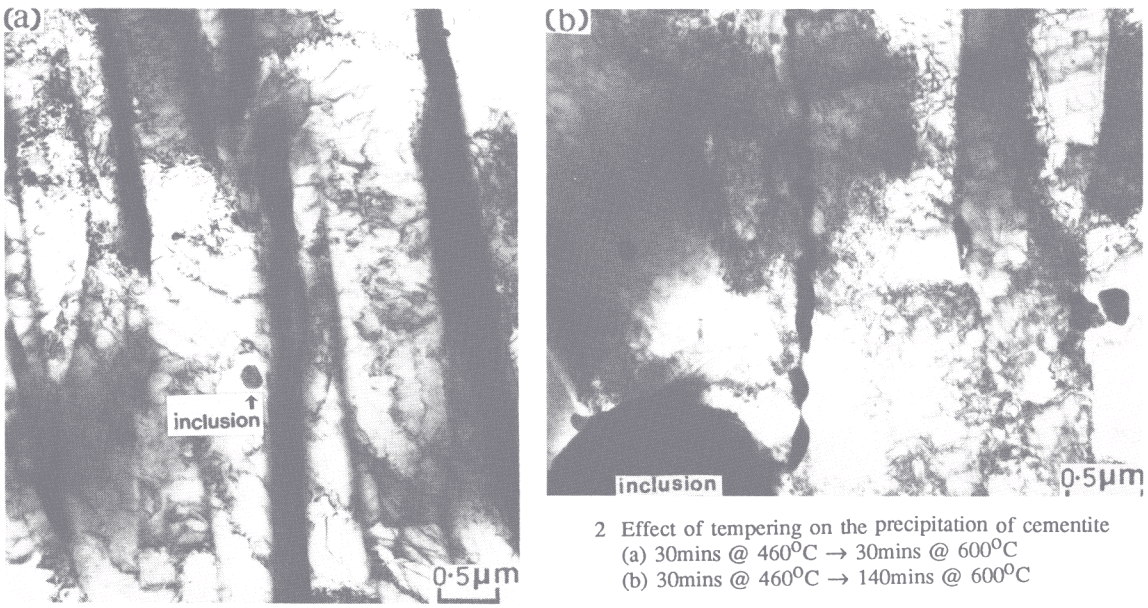
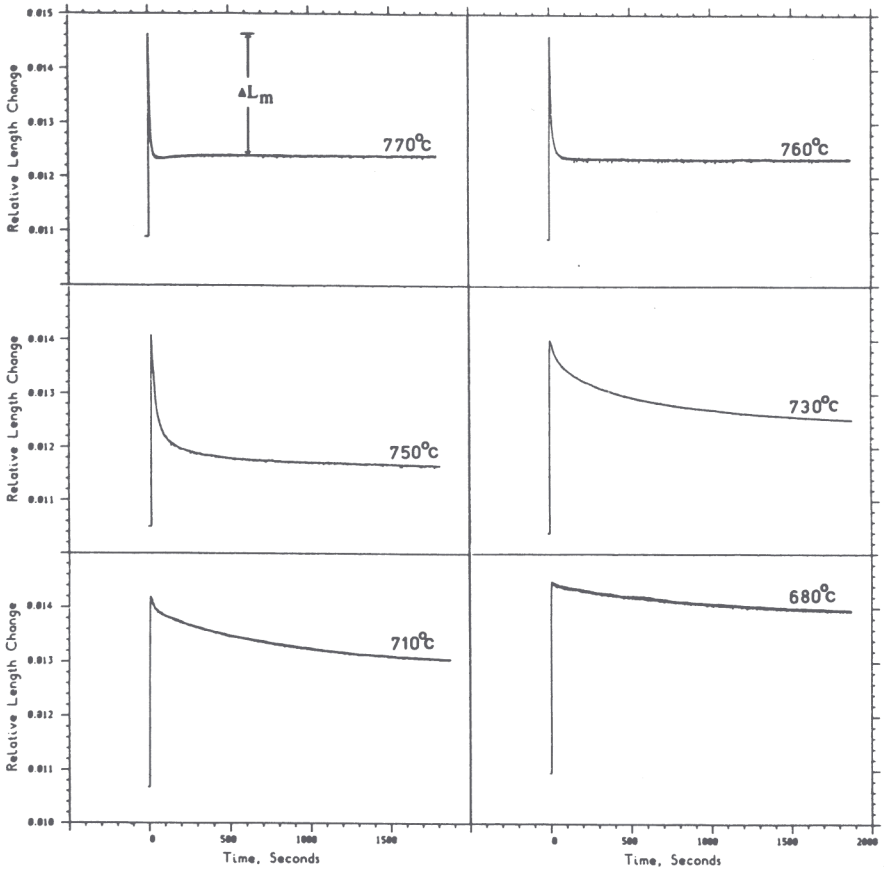


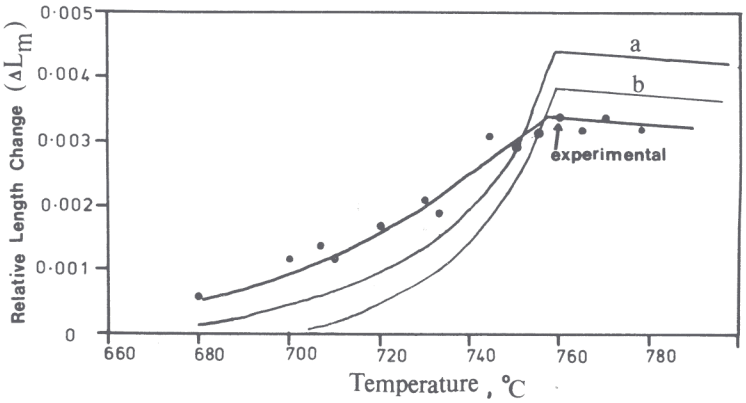
1 Bainite obtained by isothermal Transformation for 30mins @ 460°C.

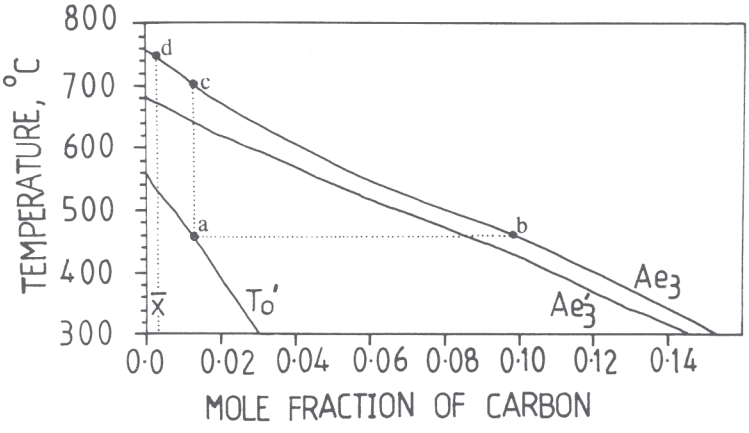


2 Effect of tempering on the precipitation of cementite
(a) 30mins @ 460°C → 30mins @ 600°C
(b) 30mins @ 460°C → 140mins @ 600°C



3 Isothermal reaustenitisation using dilatometry.



4 Plot of ΔL_m versus isothermal reaustenitisation temperature. The calculated curves **a** and **b** assume that the α_b reaction stops at the T_o curve or T_o' curve of the phase diagram, respectively.

5 Calculated phase diagram for the alloy used [1]