

# Bainite transformation in heterogeneous steels

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## SYNOPSIS

The effect of chemical segregation in the austenite on its transformation to bainite is investigated. It is found that the maximum volume fraction of bainite formed at any temperature is different for heterogeneous and homogeneous samples. A model has been developed, which qualitatively explains the results, if the carbon that is partitioned from bainite remains non-uniformly distributed within the austenite for the time scale of the experiment. Evidence confirming a non-uniform distribution of carbon in the residual austenite is also presented.

## INTRODUCTION

Most commercial alloys exhibit some degree of chemical heterogeneity, the extent depending on the thermomechanical history experienced by the alloy, and on the alloy composition itself. Chemical segregation can influence phase transformation behaviour, often giving rise to "banded" microstructures [1]. The banding is usually most severe for heavily alloyed materials, where the amplitude of composition variations tends to be larger.

This work is part of a study concerned with the development of theory for phase transformations in chemically heterogeneous materials. Its aim is to assess quantitatively the effect of alloying element segregation on the bainite transformation. The work reported here is conducted using 300M high strength steel, which for reasons of hardenability, strength and toughness is heavily alloyed and also contains a high level of silicon. The steel is normally used in a quenched and tempered condition but many attempts have been made to obtain good properties by isothermal transformation to upper or lower bainite [2]. Because of its high silicon concentration, the precipitation of cementite during the formation of upper bainite is suppressed, the microstructure obtained after appropriate isothermal transformation (at temperature  $T_i$ ) being a mixture of just upper bainitic ferrite and carbon-enriched residual austenite; some of the latter phase usually transforms to martensite on cooling to ambient temperature [3], but that which does not is called retained austenite. The high hardenability of the alloy means that the bainite transformation is well separated from other reactions; this, and the absence of cementite in the upper bainite, makes the theoretical analysis of the results less difficult, since at the end of the bainite transformation, almost all of the carbon is in the residual austenite.

## EXPERIMENTAL METHOD

The 300M steel was obtained as a 10cm diameter bar; its average chemical composition is Fe-0.44C-1.74Si-0.67Mn-1.85Ni-0.83Cr-0.39Mo-0.09V wt.%. Samples of the alloy were homogenised by holding at 1300°C for three days, with the specimens sealed in quartz tubes containing argon. Microanalysis experiments were carried out using an energy dispersive X-ray analysis facility (interaction volume  $\approx 4.5 \mu\text{m}^3$ ) on a scanning electron microscope. Dilatometric experiments using nickel plated 3mm diameter rods (length  $\approx 2\text{cm}$ ), were carried out

using a Theta Industries high speed dilatometer which has been described elsewhere [4]. The machine is equipped with a microcomputer capable of collecting length, temperature and time data at millisecond intervals.

## RESULTS AND DISCUSSION

In steels where the bainite transformation can develop without interference from other reactions, it is known to exhibit an incomplete-reaction phenomenon [5]; in this, the transformation stops prematurely, well before the carbon concentration of the residual austenite ( $x_\gamma$ ) reaches its equilibrium or paraequilibrium concentration. The transformation stops when  $x_\gamma$  reaches approximately a concentration ( $x_{\gamma^*}$ ) given by the  $T_0$  curve on the phase diagram. The  $T_0$  curve on a plot of temperature versus carbon concentration represents the locus of all points where ferrite and austenite of the same composition have equal free energy. To take account of the invariant-plane strain shape change accompanying the formation of bainite, a  $T'$  curve is defined in which the free energy of ferrite is increased by 400J/mol [6]; plastic relaxation and interactions between sheaves of bainite can be expected to reduce this stored energy term so that the limiting curve considered in the present work is the  $T_0$  curve.

In 300M steel, isothermal transformation to upper bainite results in the formation of bainitic ferrite (which is essentially free of carbon) and carbon-enriched austenite. Detailed microstructural studies will be published elsewhere; all of the present experiments refer to upper bainite only. Hence, if the volume fraction of bainite is known,  $x_\gamma$  can be calculated using a mass balance equation.

Fig. 1 shows the results of dilatometric experiments, in which the maximum degree of transformation to upper bainite increases as the transformation temperature decreases; values of  $x_\gamma$  calculated using these data are compared with the computed  $T_0$  curve in Fig. 2, which confirms that in homogeneous 300M steel the transformation exhibits a classical incomplete reaction phenomenon. Consequently, the  $T_0$  curve can be used to estimate the maximum volume fraction of bainite that may form at any isothermal transformation temperature.

Microanalysis experiments indicated that the minimum ranges of concentration (since the resolution of the technique is finite) of substitutional alloying elements in the as-received alloy are approximately: 1.55-1.97Si, 0.56-0.89Mn, 1.74-2.06Ni, 0.35-0.88Mo, 0.77-1.44Cr, 0.0-0.18V, wt.%. Because of its high diffusivity, it is expected that the activity of carbon is uniform before transformation. Fig. 3 compares the optical microstructures of the homogeneous and heterogeneous samples after isothermal transformation to bainite (until reaction ceases), the latter showing clear evidence of banding, with the bainite restricted to regions containing relatively low solute concentrations. Both Fig. 1 and Fig. 3 show that the maximum degree of transformation to bainite is higher in the homogeneous than in the heterogeneous samples. The results differ from those for allotriomorphic ferrite [7], since in that case, when the volume fraction of transformation is below 0.5, and if the carbon is distributed homogeneously throughout the residual  $\gamma$ , the extent of reaction is higher for heterogeneous alloys. For bainite on the other hand, the lower transformation temperatures

and the fact that carbon is often trapped in isolated regions of  $\gamma$  (between subunits) [8] could cause much of the carbon to remain in the close proximity of the bainite. For the heterogeneous alloys where, unlike the homogeneous alloy, bainite only nucleates in the solute-depleted regions (Fig. 3), the solute-rich regions of the sample do not transform, and indeed may not act as a site for the dumping of excess carbon (this would allow further transformation in the solute-depleted regions). This would explain why the homogeneous alloys transform to a greater extent.

To model the transformation behaviour quantitatively, the heterogeneous alloy was divided notionally into slices, each of homogeneous composition. A unique composition was assigned to each slice, on the basis of composition scans on the as-received 300M steel. The averaged composition of all the slices reflected that of the steel. A  $T_0$  curve was computed for each slice and a computer algorithm was designed for the calculation of transformation behaviour for any given temperature. In this, a small but finite degree of transformation is allowed to proceed in all slices in which  $x_\gamma < x_{T_0}$ . Since transformation eventually involves the redistribution of carbon into the austenite, a new value of  $x_\gamma$  is then calculated for each slice, and if this is greater than the corresponding value of  $x_{T_0}$ , transformation is discontinued for that slice. The program stops when all slices have  $x_\gamma = x_{T_0}$ .

In one version of the model, carbon was not allowed to redistribute between slices (which are in effect impermeable). For comparison, a second set of calculations was conducted such that at each increment of transformation, the carbon concentration is allowed to equalise in all slices. Note that this is an approximation, since in reality it is the activity of carbon which should be equalised. The results of the calculations are presented in Fig. 4; the calculations based on the impermeable slices are clearly in qualitative agreement with the experimental data whereas the other model predicts the opposite result that the degree of transformation should be higher in the heterogeneous alloy.

After isothermal transformation to bainite ceased, the dilatometer specimens were helium quenched to ambient temperature, the specimen length, temperature and time being monitored throughout. This permitted the partial transformation of residual austenite to martensite to be recorded (Fig. 4, Table 1). On directly quenching a homogeneous sample to ambient temperature, a martensite start temperature ( $M_s$ ) of 280°C was observed, and this agrees reasonably with the thermodynamically calculated  $M_s$  [9] of 276°C. On the other hand, after partial transformation to bainite, the calculated  $M_s$  temperature of the residual austenite ( $x_\gamma$  calculated from the volume fraction of bainite, assuming a homogeneous distribution of carbon in  $\gamma$ ) was always lower than that measured experimentally. This establishes the existence of a non-uniform distribution of carbon in the residual austenite, since the carbon poor regions of  $\gamma$  are expected to decompose to martensite at a higher temperature.

Although the impermeable slice model predicts that for the transformation conditions studied, the heterogeneous specimens should show a lower maximum degree of transformation, quantitative agreement with the experimental data is not satisfactory. The differences between the homogeneous and heterogeneous samples are much larger than predicted theoretically. The reason for this is not understood; it is hoped that further experimental data over a wider range of alloy compositions and temperatures will provide the basis for the further development of theory. The present experiments have been conducted at temperatures which are significantly below the bainite-start temperature of the steel. It is expected that at transformation temperatures close to (or above) the  $B_s$  temperature of the homogeneous alloy, the corresponding heterogeneous alloy should show a greater degree of transformation, since its solute-depleted regions should have a higher  $B_s$  temperature. Work is in hand to confirm this.

## CONCLUSIONS

During the transformation of 300M steel to bainite over the temperature range 320-420°C, the maximum volume fraction of bainite obtained at any temperature is lower in samples

containing chemical segregation when compared with corresponding homogeneous alloys. A model has been developed which explains the results qualitatively, as long as the carbon distribution in the austenite does not homogenise during the course of the bainite transformation. That it does not do so has been confirmed, since the  $M_s$  temperatures of partially bainitic specimens have been found to be much higher than expected.

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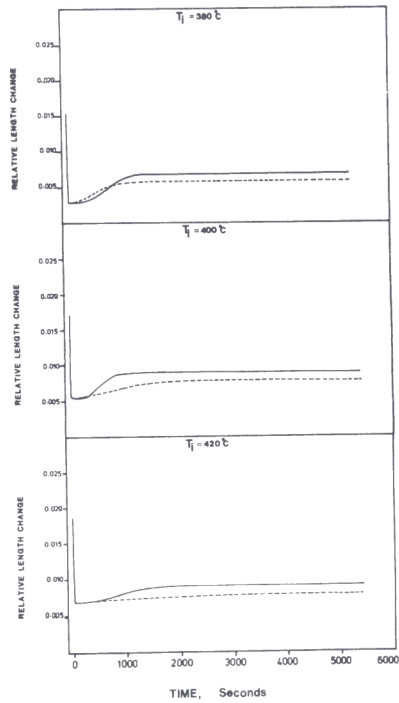
## REFERENCES

1. J. S. KIRKALDY, J. V. DESTINON-FORSTMANN and R. J. BRIGHAM, *Canad. Metall. Quart.*, 1962, 1, 59.
2. E. R. PARKER, *Metall. Trans.*, 1977, 8A, 1025.
3. R. RITCHIE, M. CEDENO, V. ZACKAY & E. R. PARKER, *Metall. Trans.*, 1978, 9A, 35.
4. J. R. YANG and H. BHADSHIA, "Welding Metallurgy of Structural Steels" Denver, ASM, in press.
5. J. W. CHRISTIAN and D. V. EDMONDS, "Phase Transformations in Ferrous Alloys", TMS-AIME, 1983, 293.
6. H. K. D. H. BHADSHIA, *Acta Metall.*, 1981, 29, 1117.
7. H. K. D. H. BHADSHIA, *Scripta Metall.*, 1983, 17, 857.
8. H. BHADSHIA and A. WAUGH, *Acta Metall.*, 1982, 30 775.
9. H. K. D. H. BHADSHIA, *Met. Sci.*, 1981, 15, 178.

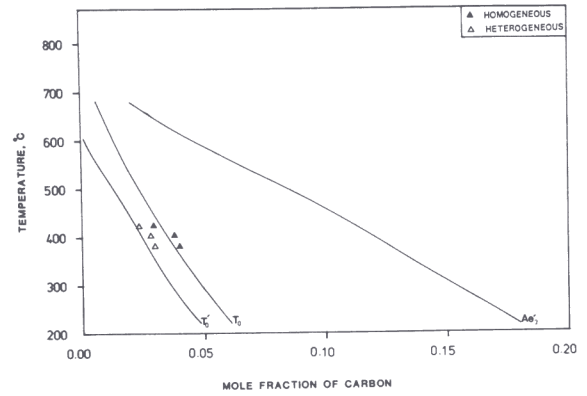
Table I: Calculated and experimental  $M_s$  temperatures for homogenised and as received (heterogeneous) samples, after partial isothermal transformation to bainite.

	$T_i, ^\circ\text{C}$	$M_s, ^\circ\text{C}$	$M_s, ^\circ\text{C}$
		Exp.	Calc.
Homogeneous	420	218	177
Heterogeneous	420	262	246
Homogeneous	400	125	63
Heterogeneous	400	230	177
Homogeneous	380	100	38
Heterogeneous	380	170	160

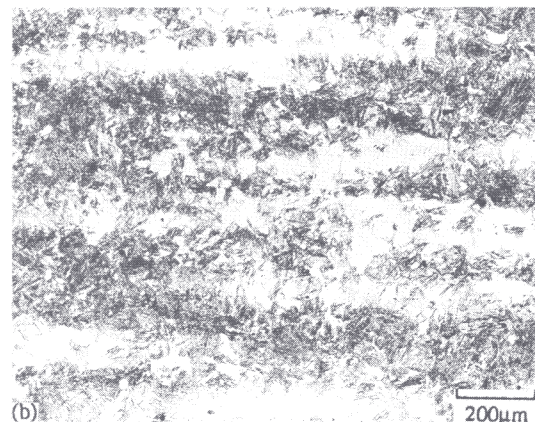
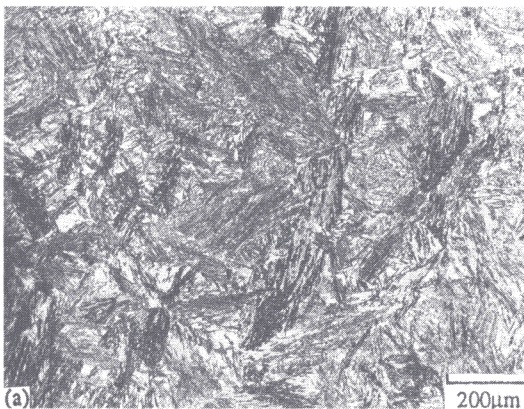




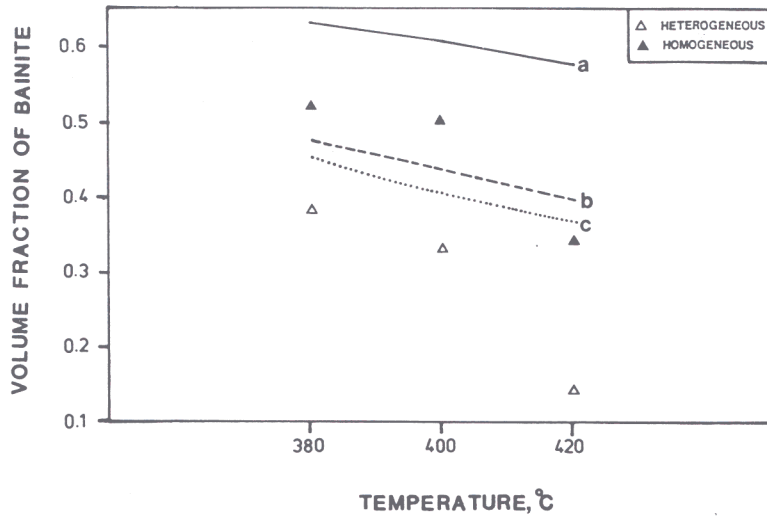
1 Dilatometric curves showing isothermal transformation to bainite. Continuous curves refer to homogeneous samples, dashed curves to corresponding heterogeneous samples.



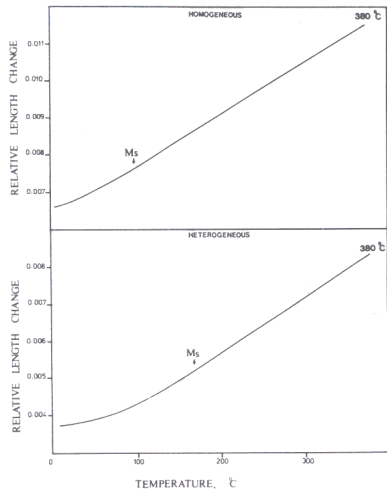
2 Calculated phase diagram for 300M steel, and experimental data indicating that the transformation to bainite stops when  $x_\gamma$  reaches the  $T_0$  curve.



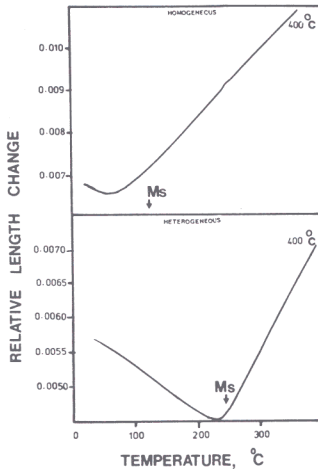
3 Optical microstructures after isothermal transformation at 400°C. (a) Homogeneous alloy, (b) heterogeneous alloy.



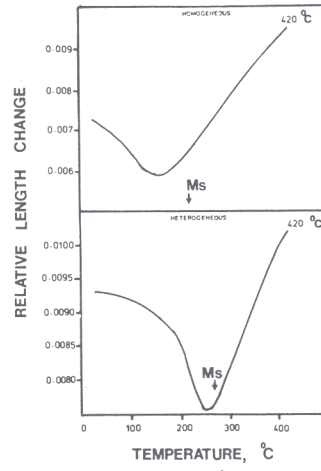
4 The curves are calculated on the basis that transformation to bainite stops when  $x_y < x_{T_0}$ . Curve b is for homogeneous samples; curve c is calculated for heterogeneous samples, using the impermeable slice model, curve a being the corresponding calculation when diffusion of carbon between slices is permitted. The points represent experimental data.



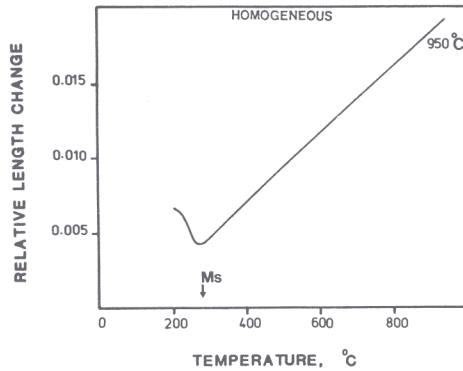
(a)



(b)



(c)



(d)

5 Dilatometric curves showing the transformation of the residual austenite during cooling from the bainite transformation temperature. (a) After isothermal holding at 380°C, (b) after isothermal holding at 400°C, (c) after isothermal holding at 420°C, (d) directly quenched to ambient temperature.