

Microstructure of lower bainite formed at large undercoolings below bainite start temperature

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Metallographic observations of steel samples transformed isothermally to lower bainite at temperatures well below the bainite start temperature have revealed that it is possible for separately nucleated platelets to coalesce during prolonged growth. The circumstances conducive to the coalescence process are discussed, and a mechanism is proposed to explain why a bimodal size distribution of plate thickness is sometimes observed for lower bainite and martensite.

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Introduction

It is often observed, though rarely highlighted or reported, that lower bainitic microstructures can contain a strongly bimodal size distribution of plates. A few coarse plates can be observed on an optical scale, whereas the remaining microstructure consists of much finer plates which can only be resolved using transmission electron microscopy.¹⁻³ This is also a feature of lath martensitic microstructures.⁴

It was necessary during the course of some experiments on the design of high strength steels, to undertake a detailed characterisation of the bainitic microstructures of a series of alloys. The experiments revealed a mechanism for the generation of the bimodal size distribution; these results are reported in the present paper.

Experimental methods

The chemical compositions of the experimental alloys are given in Table 1. The alloys were made as vacuum melts at the Swinden Laboratories of British Steel. The as cast ingots were forged and rolled down to 16 mm thick by 150 mm wide plates. The alloys are the subject of another investigation on the mechanical properties of bainitic steels, but are convenient for the present work because they all contain sufficient silicon (about 2 wt-%)* to retard the precipitation of cementite, and about 2% manganese for the purpose of hardenability. The series of alloys can be classified into three levels of carbon concentration (Table 1), which is useful because the formation of lower bainite is sensitive to the interstitial content.

Each alloy, in the form of a 10 mm dia. steel bar, was sealed in a quartz tube containing pure argon, for homogenisation at 1300°C for three days. The bar was then machined into cylindrical samples 8 mm in diameter and 12 mm in length. To ensure rapid cooling, a 3 mm dia. longitudinal hole was drilled in each cylinder. The specimens were then nickel plated using a two stage procedure, striking and plating. The striking process used a solution consisting of 250 g nickel sulphate and 27 mL concentrated sulphuric acid in 1 L distilled water, at 50°C for 3 min and a current density of 7.7 mA mm⁻². The plating was carried out at 50°C for 15 min and a current density of 0.4 mA mm⁻² in an electrolyte containing 140 g nickel sulphate, 15 g ammonium chloride, and 20 g boric acid in 1 L distilled water.

Transformation experiments were performed using a computerised thermomechanical simulator Thermec-

master-Z, equipped to simultaneously record the diametric and longitudinal strains, time, temperature, and load data. The specimen was heated using a radio frequency induction furnace, the temperature being monitored using a Pt/Pt-10%Rh wt-% thermocouple spot welded to the sample. It was ensured that the temperature variation along the length of the sample was less than 5 K.

For the alloy with only 0.095% carbon, the heat treatments could not be conducted in the thermomechanical simulator because of inadequate hardenability. The alloy was therefore heat treated using a salt bath, in order to reach the desired isothermal transformation temperature without undesired transformation occurring at elevated temperatures. After completion of the desired heat treatment in the salt bath, the samples were quenched in water at ambient temperature. In all cases, the austenitisation was carried out at 1000°C for 600 s.

Transmission electron microscopy was used to characterise the microstructures. Thin foils were prepared by electropolishing discs in a mixture of 5% perchloric acid, 25% glycerol, and 70% ethyl alcohol. The polishing potential was 55 V at a current of 20-30 mA, and at a temperature of about -10°C. The thin foils were examined using a Philips EM-400ST transmission electron microscope operated at 120 kV.

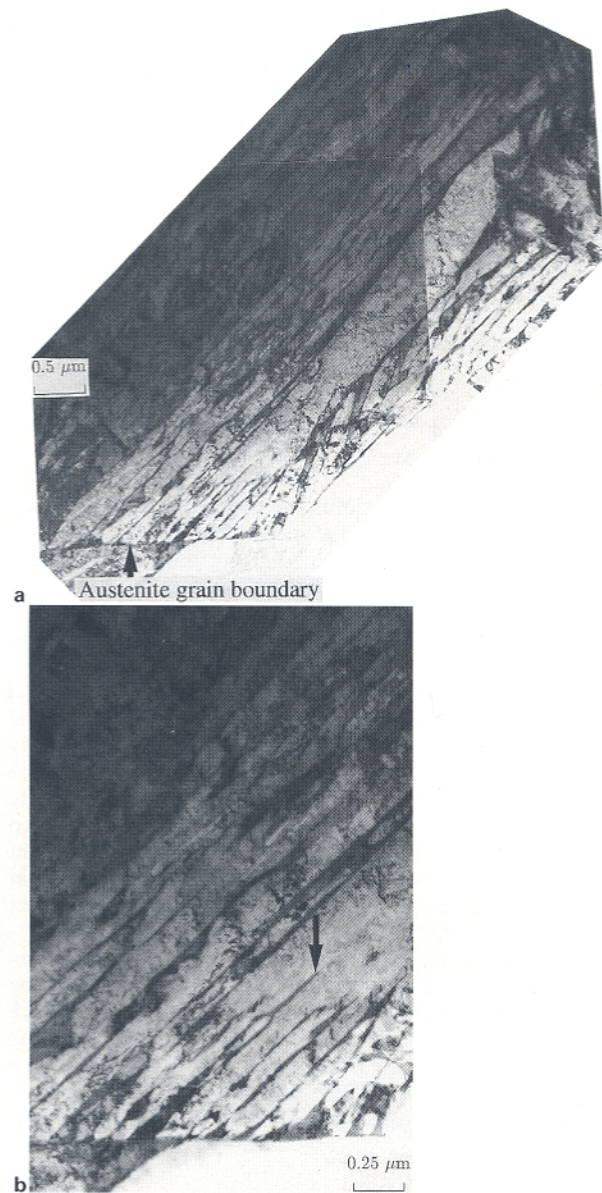
Results and discussion

Extensive routine metallography and dilatometric experiments were carried out to determine the upper and lower bainite transformation temperatures and the martensite start temperatures (B_s , LB_s , and M_s respectively). The details of the determination of the transformation temperatures are described elsewhere.⁵ It is not necessary to record the routine metallography that was used in order to determine the transformation temperatures. However, some particularly interesting observations were made during the formation of lower bainite at temperatures close to the

Table 1 Chemical compositions (wt-%) of experimental alloys: designations 'L', 'M', and 'H' refer to low, medium, and high carbon steels respectively

Alloy	C	Si	Mn	Ni	Cr
L1	0.095	1.63	1.99	...	1.97
L2	0.1	1.77	2.12	2.0	...
M1	0.27	1.98	2.18	...	1.9
M2	0.27	2.01	2.16	2.07	...
M3	0.26	1.85	2.10
M4	0.26	1.93	2.04	...	1.02
H1	0.46	2.10	2.15
H2	0.44	2.13	2.14	...	0.5

* All compositions are in wt-% unless stated otherwise.



a montage showing coalescence of subunits nucleated at austenite grain boundary, into coarse plate of lower bainite; b higher magnification image of austenite grain boundary region, showing carbides within lower bainite

1 Alloy L1, isothermally transformed at 380°C for 50 s before water quenching; arrow indicates coalescence of subunits which began their growth at austenite grain boundary located near and parallel to lower edge of micrograph (TEM)

martensite start temperature M_S ; these are discussed below.

Considering the lowest carbon alloy (L1) first, lower bainite forms during isothermal transformation at temperatures less than 390°C (Table 2), the martensite start temperature being 370°C. Figure 1a is a montage of a sample transformed isothermally at 380°C, a temperature which is just 10 K above M_S . The transformation begins with the formation of the usual small subunits of lower bainite at an austenite grain boundary (illustrated at higher magnification in Fig. 1b), but the subunits then coalesce to form a single larger plate. The coalescence is accompanied by a thinning of the austenite films between the lower bainite subunits. The films eventually vanish as the subunits develop into the coarser structure.

This striking change in the form of growth can in fact be understood when transformation occurs at a large

undercooling below the B_S temperature. There is then a large austenite grain boundary nucleation rate, giving rise to numerous adjacent subunits, the collection of subunits being the classic sheaf of bainite. Given that the adjacent subunits all have identical orientation, there is no reason why they should not combine and grow as a single unit, assuming that:

- (i) there is sufficient chemical driving force to sustain the larger amount of strain energy associated with a thicker plate
- (ii) there is nothing to stifle the lengthening of the subunits.

The first condition is satisfied because the transformation illustrated in Fig. 1 has occurred at the largest feasible undercooling below B_S (i.e. only 10 K above M_S). The chemical driving force available for diffusionless transformation at this temperature, for alloy L1, is estimated to be $\Delta G^{T^*} = -1149 \text{ J mol}^{-1}$.

This calculation was carried out using methods described elsewhere.⁶ The transformation is accompanied by an invariant-plane strain shape deformation with a shear component s of ~ 0.22 and a dilatational component directed normal to the habit plane δ of ~ 0.03 . If this is elastically accommodated then Christian⁷ has shown that the resulting molar strain energy G_S is given by

$$G_S = \frac{\mu V_m \pi c}{1 - \nu} \frac{\pi c}{4a} (\delta^2 + 0.5(2 - \nu)s^2) \dots \dots \dots (1)$$

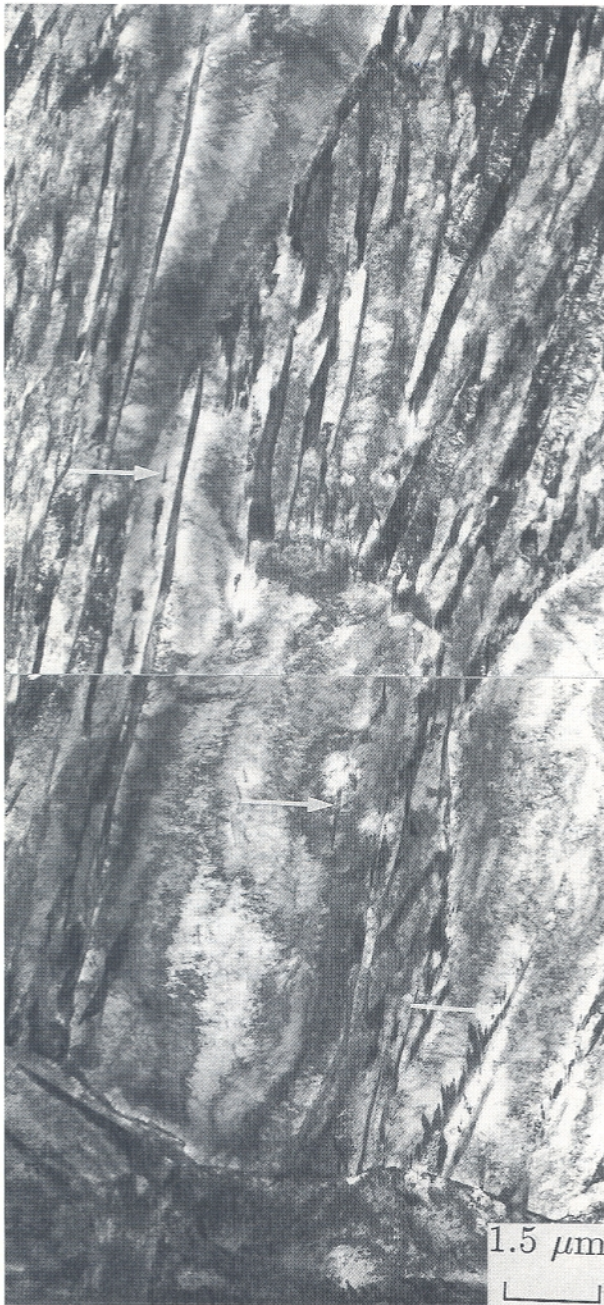
where c is the plate thickness, a is the plate length, ν is the Poisson ratio, μ is the shear modulus of austenite, and V_m is the molar volume of the austenite. Using temperature dependent elastic constants,⁸ by setting $G_S = \Delta G^{T^*}$, it can be demonstrated that for a chemical driving force of $\Delta G^{T^*} = -1149 \text{ J mol}^{-1}$, an aspect ratio of $c/a \sim 0.05$ can easily be tolerated for a plate which is elastically accommodated in austenite. The apparent aspect ratio of the plate shown in Fig. 1a, is found to be about 0.1. This estimate is designated as an 'apparent' aspect ratio because the true aspect ratio can not be estimated from the two-dimensional image shown in Fig. 1b. The value estimated from Fig. 1b is therefore likely to be an overestimate of the true aspect ratio and so is considered to be consistent with the strain energy given by equation (1).

The second condition (ii) stated above that the coalescence of subunits is only possible if the growth process can not be stifled, has several implications. It implies that such coalescence can only occur in the early stages of transformation, when the growth can not be hindered either by hard impingement with other regions of bainite, or by soft impingement processes. For example, the carbon concentration of the austenite increases as transformation progresses. The corresponding decrease in $|\Delta G^{T^*}|$ should render coalescence much less likely. It is obvious from Fig. 1a, that the coalesced plate formed first, followed by the surrounding bainite sheaves which have retained their subunit morphology.

Another reason why coalescence should only occur at relatively low temperatures is that the shape deformation accompanying the growth of bainite is not completely

Table 2 Measured transformation start temperatures, °C

Alloy	B_S	LB_S	M_S
L1	460	390	370
L2	540	390	350
M1	400	325	300
M2	460	330	300
M3	540	360	315
M4	450	340	320
H1	510	300	260
H2	450	280	245



2 Alloy L1, isothermally transformed at 380°C for 50 s before water quenching showing coalescence (arrows), and illustrating bimodal distribution of plate thicknesses (TEM)

elastically accommodated. This is the reason why subunits usually tend to reach a limiting length which is less than that of the austenite grains. Lower temperatures favour more elastic accommodation, since the yield strength of all phases increases as the temperature is reduced. Thus, the subunits can grow to longer lengths, allowing an opportunity for coalescence to occur. Figure 1a shows that in the region beside the coarse plate, the subunits are quite long in all cases where impingement has not limited growth.

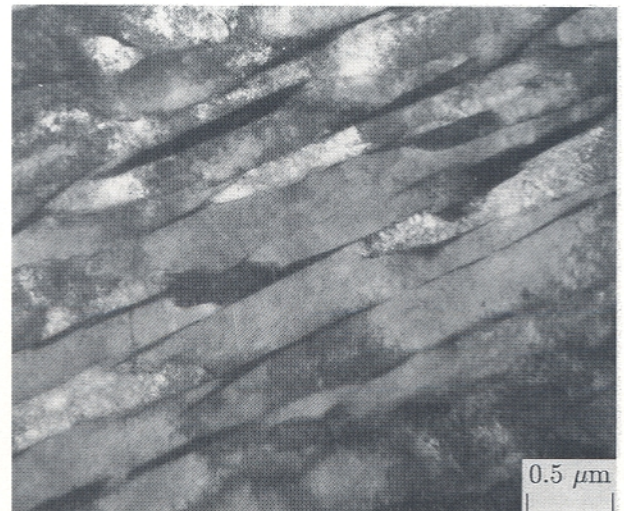
It is worth noting from Fig. 1a, that the films of retained austenite between the subunits at the austenite grain surface are much thinner than those beside the coalesced thick plate, as would be expected theoretically.⁹

Coalescence effects were frequently observed in the microstructure, which consequently appeared to contain a bimodal distribution of plates, as shown in Fig. 2.

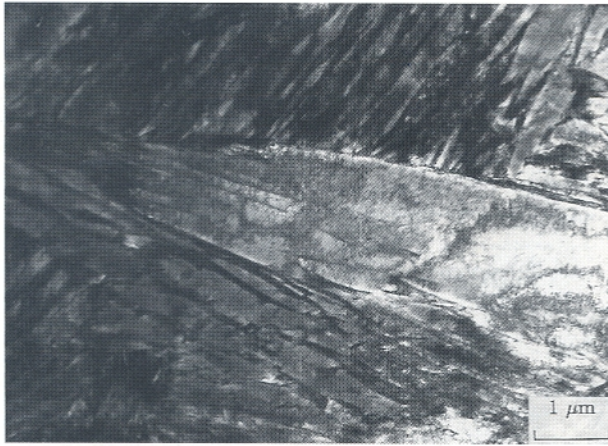


3 Alloy H1, isothermally transformed at 305°C for 150 s before quenching in thermomechanical simulator showing coalescence and intense carbide precipitation in all parts of bainite, i.e. subunits before and after coalescence (TEM)

Exactly the same observations could be made for all the other alloys. These observations were particularly striking in the high carbon alloys, where the carbide precipitation in the bainite emphasised the coalescence, as illustrated in Fig. 3 for alloy H1 transformed isothermally at 305°C.



4 Alloy L1, isothermally transformed at 400°C for 800 s before water quenching showing upper bainite with more uniform distribution of plate thicknesses (TEM)



5 Alloy M1, isothermally transformed at 325°C for 2000 s before water quenching showing subunits at tip of lower bainite plate (TEM)

Similar coalescence effects could not be found for upper bainite, presumably because of the lower chemical driving force available at higher transformation temperatures, a greater degree of plastic accommodation of the shape deformation (due to reduced yield strength), and because the carbon would tend to partition more rapidly from supersaturated ferrite at elevated temperatures. Consequently, the distribution of plates was observed to be much more uniform, as illustrated in Fig. 4.

Figure 5 shows a different observation: the tapering of the plate indicates a growing tip that is not associated with an austenite grain boundary (unlike the case illustrated in Fig. 1). The thick plate of lower bainite in alloy M1 (Fig. 5) thus appears to break up into a series of subunits at its tip. Similar observations have been reported for martensite plates;⁴ such splitting is thought to occur in order to reduce the strain energy at the growing tip of the plate when the growing tip encounters an increasing degree of constraint.

Conclusions

It is found that the fine platelets of lower bainite that nucleate at austenite grain boundaries during isothermal

transformation at large undercoolings below the bainite start temperature, tend to coalesce into coarser plates. The coalescence process is favoured by large undercoolings, and can only occur if impingement effects do not dominate at the early stages of growth. This explains the frequently observed bimodal distribution of plates found in bainitic and sometimes martensitic microstructures.

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References

1. H. K. D. H. BHADESHIA: PhD thesis, Cambridge University, 1979.
2. H. K. D. H. BHADESHIA and D. V. EDMONDS: *Metall. Trans.*, 1979, **10A**, 895.
3. R. PADMANABHAN and W. E. WOOD: *Mater. Sci. Eng.*, 1984, **66**, 1.
4. H. K. D. H. BHADESHIA and D. V. EDMONDS: 'Phase transformations 2', IV-4; 1979, London, Institution of Metallurgists.
5. L. C. CHANG: PhD thesis, Cambridge University, 1995.
6. H. K. D. H. BHADESHIA: *J. Mater. Sci.*, 1982, **17**, 383.
7. J. W. CHRISTIAN: *Acta Metall.*, 1958, **6**, 377.
8. H. I. AARONSON, M. G. HALL, D. M. BARNETT, and K. R. KINSMAN: *Scr. Metall.*, 1975, **9**, 705.
9. L. C. CHANG and H. K. D. H. BHADESHIA: *Mater. Sci. Technol.*, 1995, **11**, 874.