

Bainite in silicon steels: new composition–property approach Part 1

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Recent work on the mechanism of the bainite transformation has shown that the extent of transformation to bainitic ferrite, and the carbon content of the remaining austenite, can be estimated thermodynamically. This paper is concerned with the application of this recent work to the development of a quantitative relationship between the composition and some important mechanical properties of silicon containing steels, which can be transformed isothermally to an aggregate of bainitic ferrite and carbon enriched retained austenite only. It is demonstrated that the method has predictive capabilities, and forms the basis of further work (Part 2 of this study) on the theoretical design, optimization, and testing of two promising steel compositions. MS/0811a

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Most high strength steels in commercial use are heat treated either by slow cooling from the austenitizing temperature or by quenching to martensite, and subsequently tempering to obtain optimum properties. In comparison, isothermal heat treatments in the bainite transformation range have not found widespread application, partly due to the uncertainties associated with optimizing mechanical properties by such treatments, but also because of the difficulties involved in controlling the microstructures involved. The presence of undesirable, coarse carbides in conventional bainitic microstructures has tended to hinder their commercial exploitation.

Some basic studies¹⁻⁵ have established that large amounts of carbon enriched retained austenite can be found in association with bainitic ferrite in low alloy steels, particularly when the latter contain carbide inhibitors such as silicon and aluminium (>1 wt-%).

Subsequent microstructure–property work on such steels has indicated that the carbon enriched retained austenite is undesirable because of its mechanical and thermal instability: it is assumed that both these instabilities lead to the formation of hard phases (high carbon, untempered martensite, or various carbides) which can be detrimental to toughness. The presence of relatively soft austenite also seems to reduce the level at which applied stress causes general yielding.⁶

Nevertheless, the basic approach of utilizing dispersed austenite to enhance toughness should be sound, if the problem of instability can be overcome. In particular, a microstructure based on *only* bainitic ferrite and residual austenite (this work is concerned with such microstructures, obtained by the judicious use of silicon as an alloying

element) seems attractive, since the coarse carbides that accompany bainite formation in other steels can be avoided completely. The purpose of the present study was to use the results of some recent research⁷⁻¹⁰ on the mechanism of the bainite transformation to quantitatively define a relationship between the alloy composition and some important mechanical properties. The details of that research will not be discussed here but the essential points for present purposes are given below.

The bainite transformation is a diffusionless transformation in which the growth of a sheaf occurs by the martensitic propagation of subunits: the redistribution of carbon from these subunits to the residual austenite occurs *after* the actual transformation has taken place. This means that the maximum extent of transformation to bainite thermodynamically permissible (in the absence of any interference from other transformations) is given by the condition that the free energy change for diffusionless transformation be negative, after allowing for the stored energy of bainite (other requirements are discussed in Ref. 4). This condition is approximately equivalent to stating that the carbon content of the residual austenite must not exceed the T_0 line on the phase diagram (the T_0 line is the locus of points where the austenite and ferrite free energy curves intersect, *see* Ref. 9); it has been experimentally established that *on average* the carbon content of the residual austenite approximately conforms with the T_0 line. This allows the thermodynamic prediction of the maximum extent of isothermal transformation to bainitic ferrite from a knowledge of the steel composition only.

In this paper (Part 1 of the study) the steel (Fe–0.43C–3.0Mn–2.02Si, wt-%) used to develop the structure–property model is the one that has been studied extensively in previous investigations^{7,9,10} on phase transformations. Part 2 (following) is concerned with the application and extension of the concepts developed in Part 1 towards the prediction of the compositions of two usable and economical steels (based on the austenite–bainitic ferrite microstructure) which will be shown as having advantages comparable to those of tempered-martensite based steels of equivalent strength. Throughout this work, every attempt has been made to make the composition–property model as simple and easy to use as possible, without compromising any of the essential arguments involved.

EXPERIMENTAL PROCEDURE

The chemical compositions (wt-%) of the three alloys studied are given in Table 1. Alloy A was prepared as a 20 kg vacuum induction melt from high purity base materials. The ingot was forged and hot rolled to 10 mm dia. rod, followed by a hot-swage reduction to 8 mm dia. rod. The alloys B and C were prepared as 65 g melts from pure constituents in an argon arc furnace. The ingots were sealed in quartz capsules (under a partial pressure of argon) and homogenized for three days at 1250°C, before hot reduction to 3 mm dia. rods.

Quantitative metallography was carried out on a Quantimet image analysing computer. The electron micros-

Table 1 Chemical composition alloys studied, wt-%

Alloy	C	Mn	Si	S	P	Sb	As	Cu	Cr	Ni
A	0.43	3.00	2.02	0.002	0.002	0.001	0.002	0.004	0.01	0.01
B	0.19	2.90	2.13
C	1.10	3.00	2.02

copy and X-ray analysis procedures used have been detailed elsewhere.¹¹

The impact specimens used were 8 mm in diameter and 45 mm long, with a notch 1 mm deep machined (before any heat treatment) with a standard Hounsfield notching tool, with the specimen supported in a jig with an eccentricity of 35 mm. Hot impact testing was facilitated *in situ* by jets of furnace heated air directed at the specimen. The continuous monitoring of temperature was made possible by means of a thermocouple attached directly to the specimen (in the vicinity of the notch).

During austenitizing heat treatments, the specimens were protected by both a dynamic argon atmosphere and a proprietary coating compound. Isothermal transformations were conducted in a molten tin bath, covered with a layer of active charcoal.

RESULTS AND DISCUSSION

Hardness analysis

The strength, as a function of the volume fraction of bainitic ferrite, retained austenite, and martensite was investigated with the help of hardness (HV) tests. Electron microscopy was used to confirm that the transformation product consisted of only bainitic ferrite, retained austenite, and some martensite following isothermal transformation at 363°C. As expected (see Ref. 11), isothermal transformation at 270°C additionally resulted in some fine cementite within the bainitic ferrite. These microstructures have been extensively researched¹¹ and will be referred to as upper and lower bainite, respectively, for reasons given in Ref. 11. Typical images of these microstructures are given in Fig. 6 (upper bainite) and Figs. 10 and 11 (lower bainite) of Ref. 11.

The hardness contributions of the individual constituents could not be determined directly because of the fine scale of the microstructure. The contribution of the retained austenite was deduced from the hardness of the Fe-3Mn-2.02Si-1.1C alloy, after austenitizing and water quenching (WQ). This treatment resulted in a specimen which was almost completely austenitic and, by carefully avoiding the few plates of martensite that were present, the hardness of the austenite was found to be 240 HV: this was taken to be representative of the retained austenite associated with the microstructures discussed above even though the carbon

content of the austenite varies as a function of the volume fraction of transformation to bainitic ferrite. However, the strength of austenite is not sensitive to its carbon content.¹⁰

The variation of the carbon content of the austenite x_γ as a function of the average carbon content of the alloy \bar{x} , and the volume fraction of bainitic ferrite V_b is given by⁷

$$x_\gamma = \bar{x} + V_b(\bar{x} - s)/(1 - V_b) \quad (1)$$

where s is the amount of carbon trapped in the bainitic ferrite, either in solid solution ($s = 0.03$ wt-%), or in the form of carbides ($s = 0.27$ wt-% for the 270°C heat treatment¹¹).

The hardness of the martensite in alloy A (directly water quenched following austenitization at 1080°C for 10 min) was found to be 620–660 HV, but this cannot represent the hardness of the martensite that forms from the *residual austenite*, not only because of the carbon enrichment (equation (1)) but also because direct quenching gave a lower martensite hardness compared with austenite which first had been quenched to a lower temperature (above M_s) and then water quenched. The latter type of heat treatment is representative of the conditions to which the austenite is exposed following isothermal transformation to bainite. This effect was allowed for by step quenching a series of specimens of alloy A from the austenitizing temperature to between 370 and 270°C ($M_s = 220^\circ\text{C}$, Ref. 11), holding for short periods so that any isothermal transformation was limited to $V_b < 0.005$ (too small to affect the martensite hardness, see equation (1)), and immediately water quenching. After allowing for the levels of retained austenite, which were determined by X-ray techniques, the martensite hardness was found to be 795 HV.

The difference between the step quenched and directly quenched values of hardness can be attributed to the relative ability of the two treatments to eliminate thermal gradients within the specimens: such effects may lead to variations in the degree of any kind of autotempering in the martensite. A hardness of 795 HV was taken to be representative of martensite (with a base composition of about 3Mn-2Si-Fe) containing 0.43 wt-%C, since all the specimens under investigation had effectively been step quenched to the isothermal transformation temperature and then water quenched.

Table 2 Quantitative analysis of microstructure of specimens of alloy A

Heat treatment	V_b	V_γ	V_α	Hardness, HV50
363°C for 2015 min	0.46	0.27	0.27	433
1070 min	0.52	0.33	0.15	420
160 min	0.39	0.27	0.34	428
80 min	0.48	0.27	0.25	492
20 min	0.02	0.11	0.87	750
10 min	0.005	0.085	0.91	759
5 min	0.005	0.10	0.895	710
270°C for 6060 min	0.75	0.15	0.10	516
1730 min	0.64	0.15	0.21	492
160 min	0.63	0.14	0.23	554
80 min	0.22	0.10	0.68	590
23 min	0.005	0.065	0.935	754
5 min	0.045	0.080	0.875	756

The variation of the martensite hardness with carbon content (1020 HV/wt-% change in carbon concentration) was determined by measuring the hardness of martensite in alloy B (560 HV), using the step-quenching technique. Before the isothermal transformations listed in Table 2 the specimens were austenitized at 1200°C for 7 min. A layer, about 0.5 mm thick, was removed from the specimen surface which was then chemically polished before testing. The hardness tests were conducted before hot mounting the specimens in plastic for metallographic work.

The data of Table 2 were analysed in terms of a simple model, in which the total hardness H is expressed as a linear combination of the contributions from the individual components, such that

$$H = V_\gamma H_\gamma + V_b H_b + V_\alpha H_\alpha \dots \dots \dots (2)$$

where the martensite hardness H_α must be a function of the carbon content of the residual austenite, such that

$$H_\alpha = Q V_b (\bar{x} - s) / (1 - V_b) + H_0 \dots \dots \dots (3)$$

where

Q = the dependence of martensite hardness on carbon content, found to be 1020 HV/wt-% of carbon in solid solution

H_0 = the martensite hardness at a carbon content of \bar{x} (795 HV in the present work)

V_γ = the volume fraction of retained austenite

V_α = the volume fraction of martensite

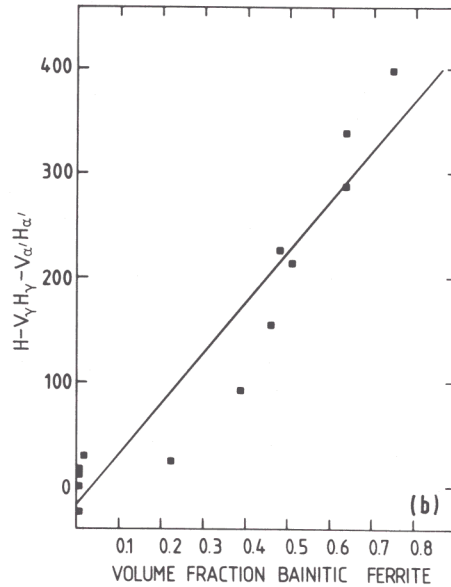
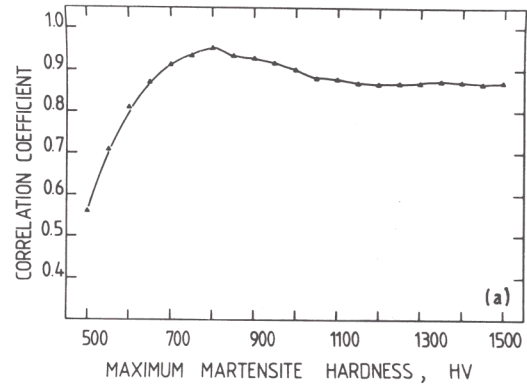
H_b = the hardness of bainitic ferrite.

In considering the deformation of an aggregate of hard and soft phases, it is necessary to allow for the fact that any increase in the hardness level of the martensite beyond an unknown level N should have no effect on the hardness of the microstructure as a whole. The martensite would then behave as an elastic constituent, the general plastic flow being accommodated by the softer constituents. Under these circumstances, any increase in the martensite hardness should have little effect on the observed macroscopic hardness. The value of N appropriate to the present experiments was deduced as indicated below.

Equation (2) shows that a plot of $(H - V_\gamma H_\gamma - V_\alpha H_\alpha)$ v. V_b would have a slope of H_b . The data of Table 2 were plotted in this way, but a limit N was placed on the maximum hardness that the martensite may achieve, as given by equation (3). N was systematically varied (between 1500 and 500 HV) until the correlation coefficient for the above plot was maximized (Fig. 1a), giving the best fit values of $N = 800$ HV and $H_b = 480$ HV (the relevant plot using these values is given in Fig. 1b). This gives a bainitic ferrite yield stress of about 1440 MN m⁻², which seems rather high in comparison with low alloy steel martensite (about 1000 MN m⁻²) of the same carbon (<0.03 wt-%) and manganese levels^{12,13} after allowing for the effect of silicon.¹⁴ On the other hand, the subunits of bainitic ferrite involved in the present study are separated by relatively thick regions of austenite, so that the transfer of slip between these subunits would be expected to be a complex process: in low alloy steel martensites the packets of platelets are usually separated by low-angle boundaries, so that the grain refining effect is reduced. In any case, it is evident from Fig. 1b that an analysis based on equation (2), i.e. a law of mixtures, is not strictly justified. Nevertheless, the model does give approximate information on the role of the very different constituents of the microstructure involved.

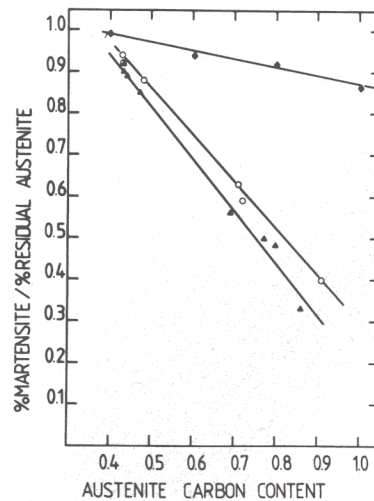
Stability of residual austenite

The amount of residual austenite at the isothermal transformation temperature is given by $(1 - V_b)$. The fraction of this quantity, which decomposes to martensite on quenching to ambient temperature, is calculable since the quantity of retained austenite can be determined by X-ray



1 a plot of linear correlation coefficient relating $(H - V_\gamma H_\gamma - V_\alpha H_\alpha)$ and N v. N ; b plot of $(H - V_\gamma H_\gamma - V_\alpha H_\alpha)$ v. V_b , using value of $N = 800$ HV

diffraction. Hence, the volume fraction of martensite is simply $V_\alpha = (1 - V_b) - V_\gamma$. It follows that the stability of the residual austenite (to martensitic transformation on quenching to the ambient temperature from the isothermal transformation temperature) may be described in terms of the



2 Graph illustrating instability of residual austenite (i.e. austenite remaining untransformed at isothermal transformation temperature) as function of austenite carbon content (wt-%)

Table 3 Hardness test results for alloy A

Heat treatment	Hardness, HV50
WQ + 247°C for 15 min, WQ	623
WQ + 347°C for 4 h, WQ	483
WQ + 508°C for 60 min, WQ	436
WQ + 597°C for 60 min, WQ	334
360°C for 115 min, WQ	430
295°C for 115 min, WQ	447
295°C for 115 min, WQ, 397°C for 60 min, WQ	461
295°C for 115 min, WQ, 500°C for 60 min, WQ	387
295°C for 115 min, WQ, 615°C for 79 min, WQ	319
243°C for 24 h, WQ	520
243°C for 24 h, 400°C for 60 min, WQ	520

parameter $V_{\alpha}/(1 - V_b)$. Figure 2 is a plot of the stability parameter as a function of the carbon content of the austenite using data from Table 2: the data for plain carbon steels are from Ref. 15.

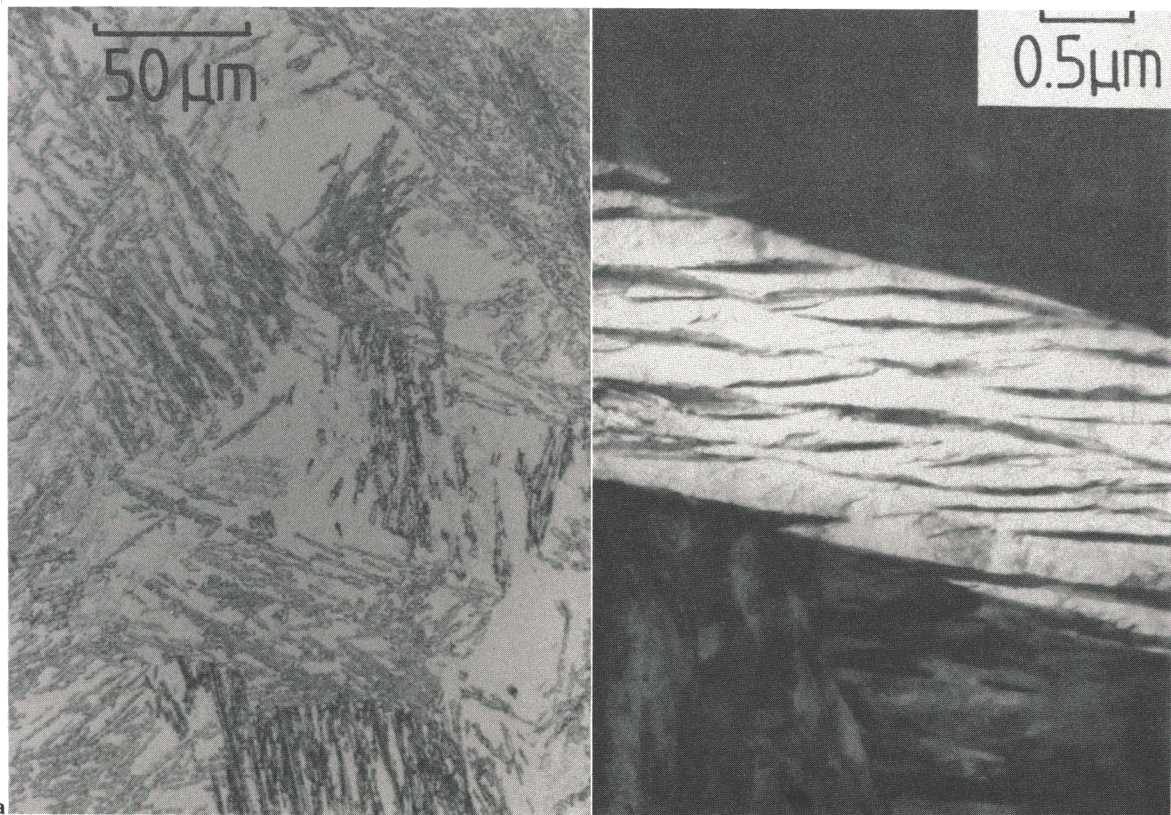
The higher stability of the austenite involved in the present experiments (relative to the plain carbon steel data) cannot be attributed simply to the difference in substitutional alloying element levels (which would mainly lead to vertical shifts in the curves of Fig. 2). It is believed that the enhanced stability is a result of the finely divided state of the austenite remaining after partial transformation to bainite; much of the plain carbon steel data is based on material containing large chunks of austenite.

Retained austenite and impact properties

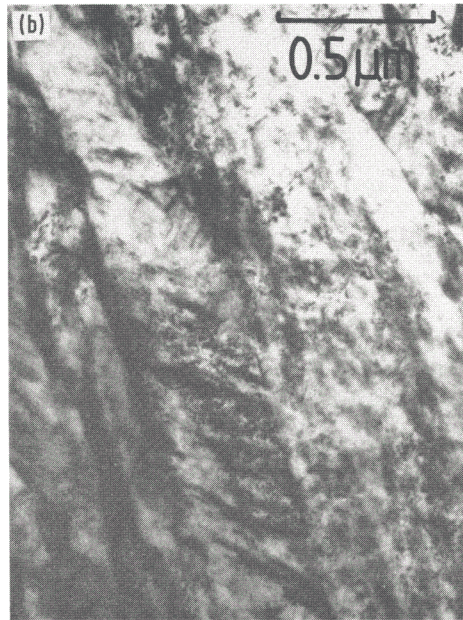
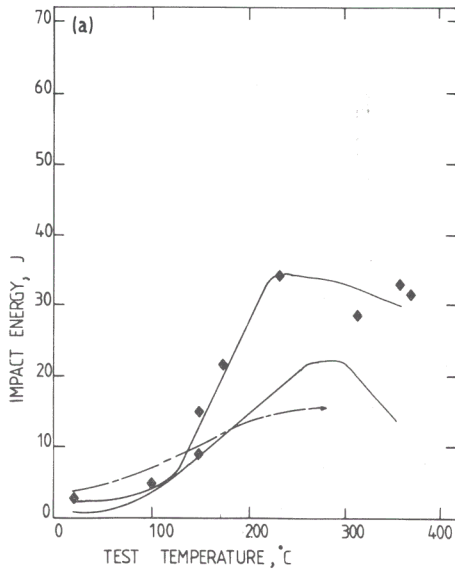
The results of hardness tests on the specimens are presented in Table 3 with the various heat treatments. Each hardness value is averaged over 10 tests: the test results are discussed below.

Isothermal transformation of alloy A at 360°C for 115 min gave a microstructure consisting of $V_b = 0.42$, $V_{\gamma} = 0.27$, and $V_{\alpha} = 0.31$. The low value of V_b indicates that much of the residual austenite present at the isothermal transformation temperature has a blocky morphology (Fig. 3a), bounded by crystallographic variants of bainite sheaves. In discussing the morphology of the austenite remaining after partial transformation to bainite, it is necessary to distinguish between the blocky morphology (exhibiting triangular shape in two dimensional sections) described above and the films of austenite which are retained between the subunits within a given sheaf of bainite (Fig. 3b, and more detailed information is given in Figs. 1b and c of Ref. 7). The carbon content of this austenite is expected to be low (about 0.75 wt-%, equation (1)), and only half of the residual austenite should be retained on quenching to ambient temperature (Fig. 2): this is confirmed by the volume fractions of retained austenite and martensite quoted above. It should be noted that the duration of the heat treatment used ensures the formation of the thermodynamically maximum amount of bainitic ferrite.

The impact properties obtained from the above microstructure are presented in Fig. 4a, with those of martensite obtained by direct quenching from the austenitizing temperature (from Ref. 16). The curves indicate that it is the large amount of martensite in the microstructure that determines the toughness: the peak in the untempered martensite curve at high test temperatures has been shown¹⁶ to correspond to the precipitation of ϵ carbide in the impact specimens during the course of the high temperature impact test. These carbides act to nucleate voids and lead to poorer impact properties. A significant drop in the upper shelf energy of the bainite curve was attributed to the *in situ* tempering of the martensite component of the microstructure, and this was verified by electron microscopic examination (Fig. 4b) of a thin foil machined from an impact



3 *a* optical micrograph showing large blocky regions of austenite in microstructure obtained by isothermal transformation at 360°C for 115 min; *b* electron micrograph illustrating bainite sheaf, with films of retained austenite separating bainite subunits



◆ 360°C upper bainite — virgin martensite — tempered martensite (245°C for 15 min)
4 a impact transition curves of as-quenched (virgin) martensite, tempered martensite, and bainite (obtained by isothermal transformation at 360°C for 115 min); **b** electron micrograph showing carbide containing martensite in specimen tested in 'fall-off' part of upper shelf region of 360°C bainite impact curve of Fig. 4a

specimen from the fall-off region indicated in Fig. 4a. The results also indicate that it is the high carbon, untempered martensite which in fact determines the toughness of the microstructure developed by isothermal transformation at 360°C.

The detrimental effect of the martensite (discussed above) can ultimately be attributed to the instability of the blocky residual austenite: if V_b could be increased, the volume fraction of such austenite would decrease, and its carbon content would increase, leading to a reduced final $V_{\alpha'}$. However, the maximum V_b obtainable at any isothermal transformation temperature is restricted by the incomplete reaction phenomenon,⁷⁻¹¹ and the only way to increase V_b would be to lower the isothermal transformation temperature (see Refs. 7 and 9). Specimens were, therefore, isothermally transformed at 295°C for 115 min, giving an upper bainitic microstructure composed of $V_b = 0.68$, $V_{\gamma} = 0.22$, and $V_{\alpha'} = 0.10$. The hardness of this microstructure was found to be 447 HV, and due to the high degree of transformation to bainitic ferrite, a greater proportion of the retained austenite was present as films between subunits. The blocky morphology was still present, but was well dispersed and on a finer scale (Fig. 5).

By examining electron micrographs (e.g. Fig. 3b) it can be estimated that about 15% of the volume contained within the boundaries of a bainite sheaf consists of retained austenite films interspersed with bainite subunits. Hence, the ratio of the volume fraction of the film type retained austenite $V_{\gamma-f}$ to that of the blocky type austenite $V_{\gamma-B}$ can be deduced from the following equations

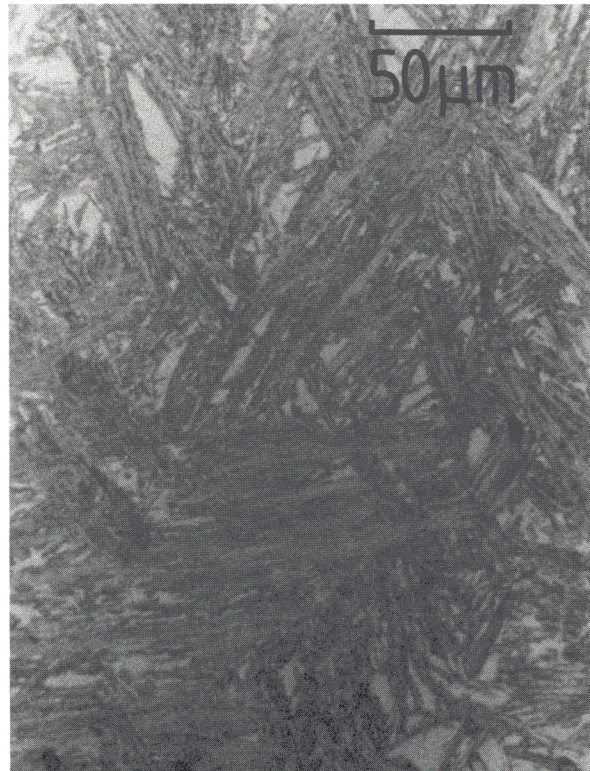
$$V_{\gamma-f}/V_{\gamma-B} = (0.15V_b)/(V_{\gamma}-0.15V_b) \dots \dots \dots (4a)$$

or alternatively

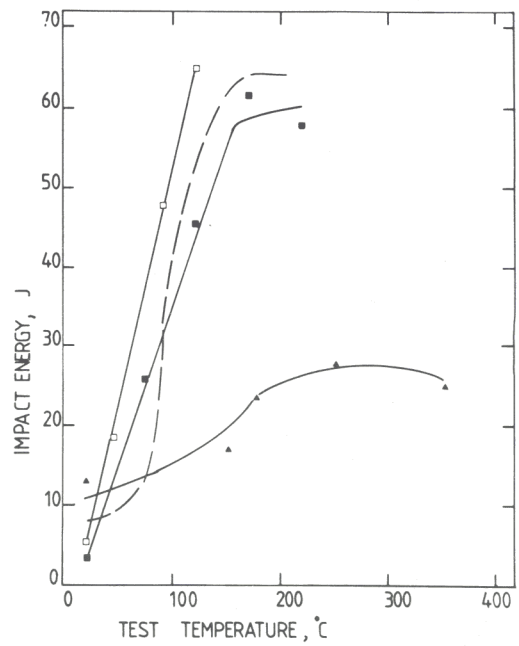
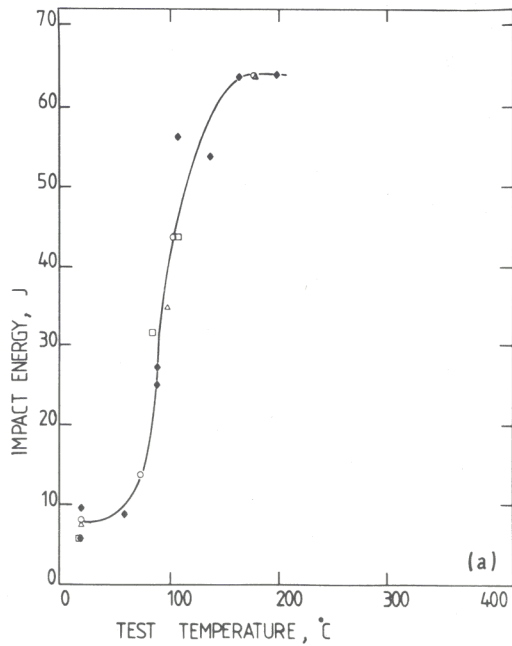
$$V_{\gamma-f}/V_{\gamma-B} = (0.15V_b)/(1-1.15V_b-V_{\alpha'}) \dots \dots \dots (4b)$$

The ratios were thus found to be 0.23 and 0.5 for the 360 and 295°C microstructure, respectively: film austenite is much more stable than blocky austenite (see Ref. 11 and results discussed below) and a large film/blocky austenite ratio should result in better toughness. The specimens isothermally transformed at 295°C showed much improved toughness (with no loss of strength) relative to the 360°C heat treatment: this toughness exceeded that of any of the

tempered martensites obtained from the same steel, despite the fact that the martensite based microstructures were often much softer (Fig. 6, Table 3). Figure 6a shows that prolonged holding at the isothermal transformation temperature (for periods longer than 115 min) has no effect on the impact properties since the maximum degree of transformation consistent with the T_0 condition has been achieved before 115 min at the isothermal temperature. The observation also indicates that little, if any, recovery

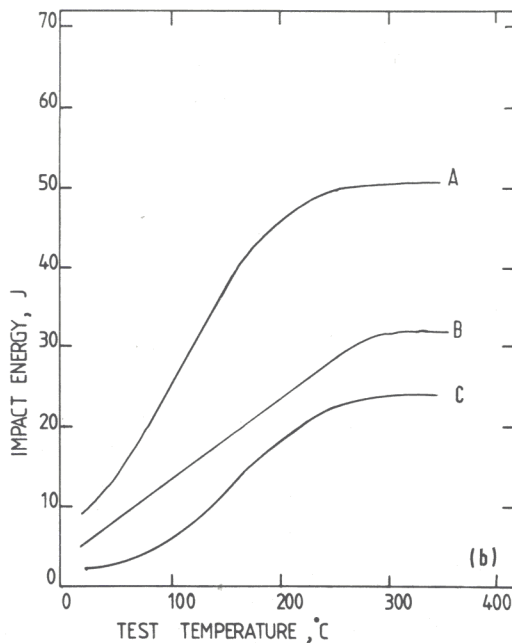


5 Optical micrograph showing more refined regions of austenite present in microstructure obtained by isothermal transformation at 295°C for 115 min



□ upper bainite tempered at 397°C for 60 min ▲ upper bainite tempered at 500°C for 60 min ■ upper bainite tempered at 615°C for 79 min

7 Impact transition curves for bainite formed by isothermal transformation at 295°C and subsequently tempered: dashed line = untempered



Time at 295°C. min

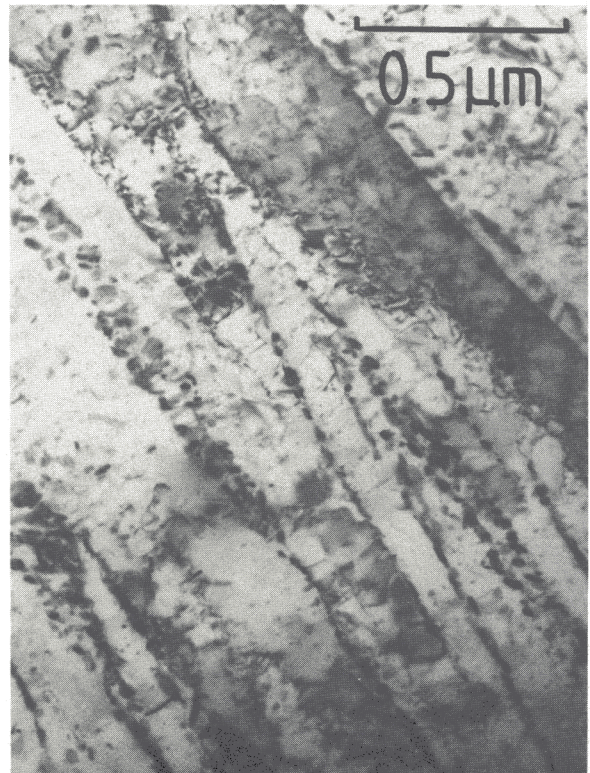
- ◆ 115
- 270
- △ 960
- 2160

6 Impact transition curves for a upper bainite formed by isothermal transformation at 295°C and b tempered martensites (reproduced from Ref. 16): A = quenched and tempered at 600°C for 60 min; B = quenched and tempered at 510°C for 60 min; C = quenched and tempered at 355°C for 240 min

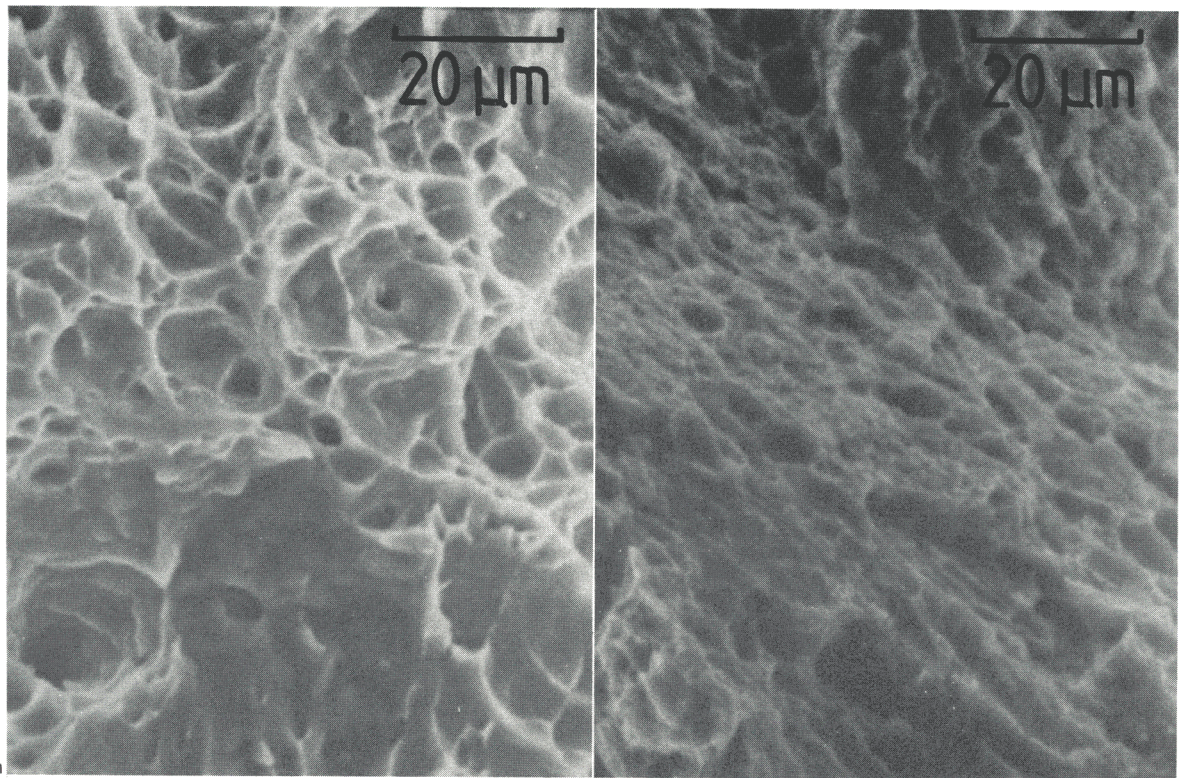
occurs at 295°C. However, tempering the above microstructure at 397°C for 60 min did result in somewhat better impact properties (Fig. 7), without any detectable change in hardness. Electron microscopy revealed that the retained austenite was still intact, although the small amount of martensite had been tempered. No significant signs of recovery were observed in the bainitic ferrite: the better toughness is therefore attributed to the tempering of the martensite present in the microstructure.

Tempering at 500°C for 60 min led to a drastic fall in the impact properties (Fig. 7), despite a drop in the hardness

value from 447 to 387 HV. Such tempering leads to the diffusional decomposition of high carbon retained austenite into bands of discretely nucleated carbides (Ref. 11 and Fig. 8). Examination of the nickel plated surfaces of impact specimens indicated that the failure was translat, and comparative fractography (Fig. 9) between the tempered and untempered specimens (from the upper shelf region of



8 Microstructure of bainite formed by isothermal transformation at 295°C, and subsequently tempered at 500°C for 60 min



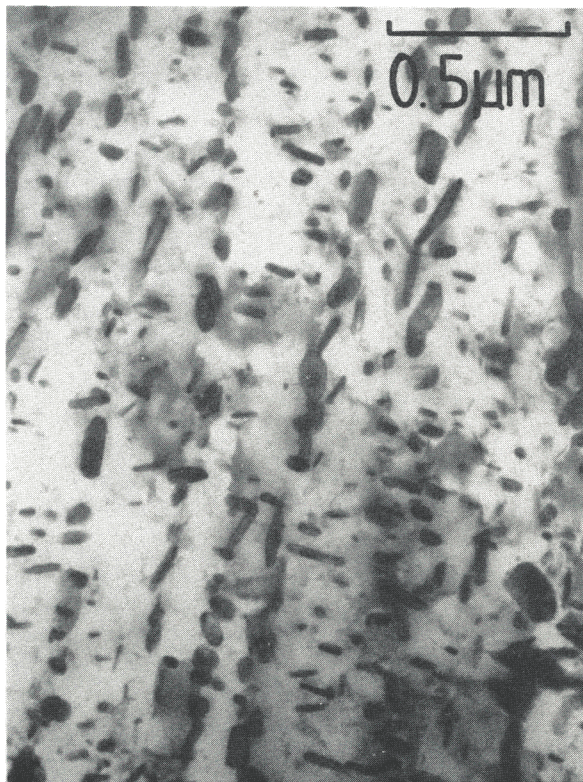
a bainite formed by isothermal transformation at 295°C; b as a but also tempered at 500°C for 60 min

9 Scanning electron fractographs from specimens fractured in upper shelf region of impact transition curves

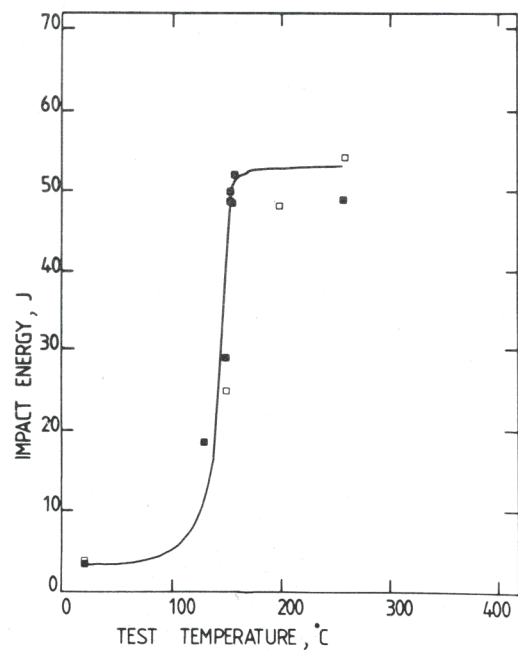
the impact transition curve) indicated the void nucleation role of the carbide particles in the former specimens. These experiments illustrate the detrimental effect of the presence of carbide particles in the microstructure. Tempering at 615°C for 80 min caused the formation of cementite par-

ticles, together with extensive recovery (Fig. 10), and a hardness of 319 HV (impact properties presented in Fig. 7).

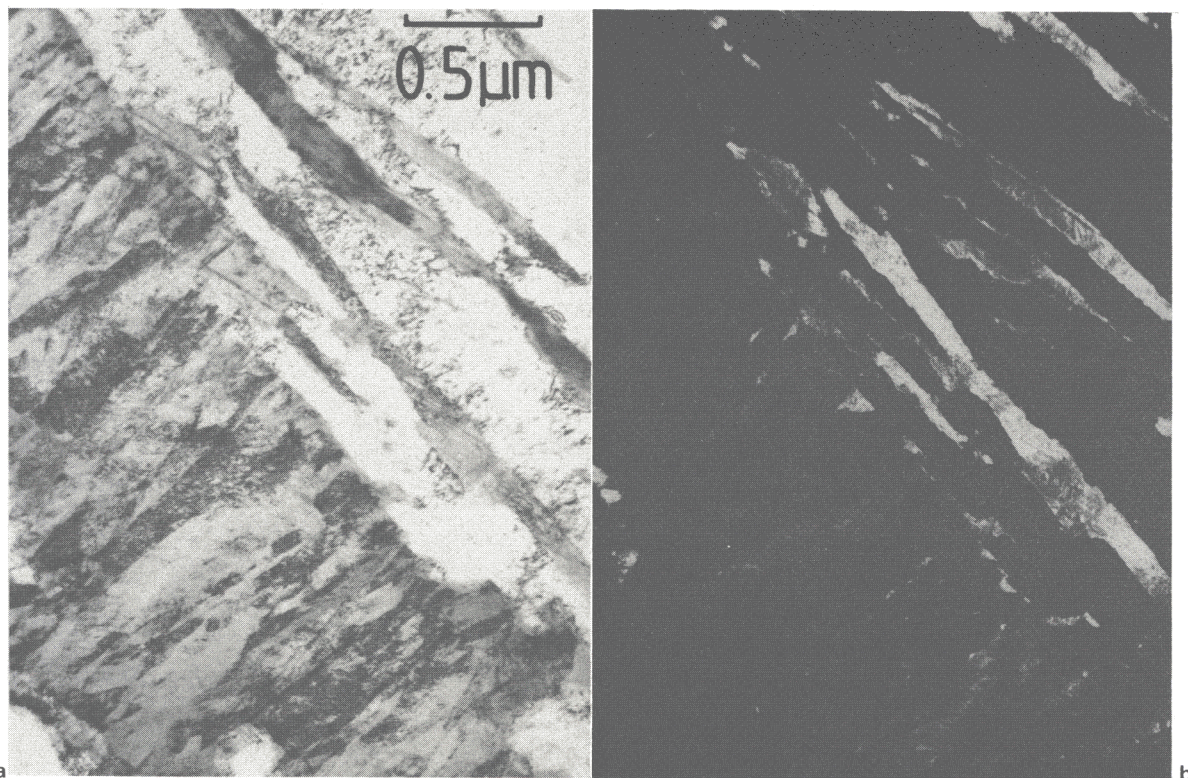
Lower bainite obtained by isothermal transformation at 243°C for 24 h was also examined (hardness = 520 HV). Transformation at such a low temperature gives a high volume fraction of transformation, and hence a decreased amount of blocky austenite together with a reduced degree of transformation to martensite on cooling to ambient temperature. The impact properties are presented in Fig. 11,



10 Microstructure of bainite formed by isothermal transformation at 295°C, and subsequently tempered at 615°C for 79 min



11 Impact transition curve for bainite formed by isothermal transformation at 243°C, in both untempered and tempered (400°C for 60 min) conditions



a bright field image; *b* retained austenite dark field image

12 Electron micrographs illustrating effect of applied stress (850 MN m^{-2}) on microstructure obtained by isothermal transformation at 295°C

and tempering at 400°C for 60 min made no significant difference to either the impact or the hardness values. This is consistent with the observations by TEM (transmission electron microscopy) that the retained austenite was unaffected, and that the microstructure contained little martensite ($V_b = 0.79$, $V_\gamma = 0.11$, $V_{\alpha'} = 0.1$).

Comparison of the lower bainite impact transition curve with those of the tempered martensites (all of which have lower hardness levels, see Table 3) clearly demonstrates that the retained austenite has a strong and beneficial influence on the properties of the base microstructure. The comparison is valid since the tempered martensite microstructure is equivalent to the lower bainite microstructure, apart from the presence of considerable retained austenite in the latter. Both the microstructures contain Widmanstätten arrays of cementite precipitates, and the significance of the comparison is enhanced by the fact that the tempered martensites have lower hardness values. Finally, it is thought that the diminished shelf energies of the lower bainite compared with the 295°C upper bainite can be attributed to the presence of cementite in the former structure.

Mechanical stability of carbon enriched retained austenite

Tensile specimens (Hounsfield no. 11) were isothermally transformed to the 295°C upper bainite, water quenched, and finally annealed at 400°C for 60 min. The purpose of this treatment was to temper the martensite formed during the quench from the isothermal transformation temperature

so that it could subsequently be distinguished from that formed as a result of the application of stress.

The specimens were stressed to various levels without fracture or necking, and were then examined by TEM. Following stressing to 1000 MN m^{-2} it was found that the thin films of austenite between the bainitic ferrite subunits remained untransformed, but the irregular blocky austenite decomposed to untempered martensite (Fig. 12). (It should be noted that the specimens had been annealed at 400°C for 60 min before testing, so that the large region of untempered martensite in Fig. 12*a* is the consequence of the stress induced transformation of prior retained austenite.) This confirms that the film austenite is more stable (and hence beneficial) than the corresponding blocky morphology. Other larger tensile specimens (8 mm dia. and 60 mm gauge length) were isothermally transformed at 295°C for 115 min and then stressed to 995 MN m^{-2} without failure or necking. The exact stress histories are listed in Table 4. Impact specimens were subsequently machined from the gauge lengths and tested as indicated in Table 4. These indicated no deterioration in toughness, a result that is consistent with the observation that the presence of islands of austenite is not a controlling feature of the toughness of the 295°C treatment, and that the films of austenite remain stable under such stresses.

SUMMARY AND CONCLUSIONS

In steels containing sufficient quantities of silicon, it is possible to transform isothermally (from austenite) to a microstructure consisting of only bainitic ferrite and carbon

Table 4 Impact energies of prestressed upper bainite specimens

Stress history	Test temperature, $^\circ\text{C}$	Impact energy, J
Stressed to 995 MN m^{-2}	120	65
Cyclically stressed $0\text{--}995 \text{ MN m}^{-2}$ 5 times	120	65
Cyclically stressed $0\text{--}995 \text{ MN m}^{-2}$ 20 times	80	31

enriched retained austenite. Under favourable circumstances, such microstructures can be exploited to give improved strength and toughness properties compared to those of conventional tempered martensite based microstructures. The combination of factors which seems to account for the exceptional properties can be summarized as follows:

- (i) the bainite transformation leads to what is effectively an ultrafine grained microstructure. The platelets resulting from the displacive character of the transformation are separated by films of stable retained austenite. Such a composite microstructure is conducive to both high strength and good toughness, since a potential crack would have to traverse not only finely spaced interphase interfaces but also varying crystal structures; in addition, its motion would tend to be dampened by enhanced deformation in the softer austenite, and perhaps by the limited transformation of this austenite at stress concentrations
- (ii) the bainite transformation in a silicon containing steel can be designed to avoid the presence of any detrimental carbides, thereby making crack and void nucleation difficult
- (iii) both the bainitic ferrite and retained austenite have extensive defect structures, leading to further strengthening
- (iv) the strengthening of the ferritic component of the microstructure is not obtained by interstitial hardening (which is very detrimental to toughness) despite the high overall carbon content that can be tolerated in the base alloy
- (v) the retained austenite can be designed to be stable to high stresses ($> 1000 \text{ MN m}^{-2}$) and over a wide temperature range.

A number of restrictions apply to the achievement of optimum properties. The restrictions are thermodynamic in origin, since only a limited amount of isothermal transformation is permissible, at a given temperature. If the degree of transformation permitted happens to be small, the properties will be poor despite the retention of relatively large quantities of austenite. This is because the austenite is found to occur in two forms: blocky and film morphologies. The former originates from the geometrical partitioning of the prior austenite grain whereas the latter is confined between bainitic subunits. The stability of the blocky austenite has been found to be poor with respect to martensitic transformation, and such austenite is undesirable as far as toughness is concerned.

Using data from the 295°C upper bainite, which exhibited a relatively good toughness–strength combination, it seems that the ratio of volume fraction of the film/blocky morphology must exceed about 0.9 for optimum properties. From equation (4b) it follows that the criterion for optimum properties is that

$$(V_{\gamma-f}/V_{\gamma-B}) = V_b/(6-7.7V_b) > 0.9 \quad (5)$$

Since V_b can be thermodynamically estimated from a knowledge of the steel composition only, the toughness behaviour of silicon bainites can be predicted, in principle, without recourse to experiment. This forms the basis of Part 2 of this study, and is an exciting criterion from the point of view of theoretical steel design.

It is anticipated that an additional opposing condition must also apply: it may be necessary to have a certain (unknown) *minimum* amount of carbon in the alloy, if the films of austenite are to be thick and continuous enough to effectively buffer the bainite subunits. This becomes obvious when the situation is exaggerated by letting \bar{x} tend to zero. The austenite films initially would be expected to become shorter and more discontinuous, and ultimately vanish, leaving little resistance to the propagation of cracks across the low-angle boundaries which would then separate the bainite subunits.

Finally, the economic significance of the above bainitic microstructure is clear: it uses carbon (the cheapest and most effective austenite stabilizer) to promote the formation of what is, in effect, a duplex low alloy steel. At the same time, the detrimental aspects of carbon in steels are avoided: not only can brittle carbides be eliminated from the microstructure, but the bainitic ferrite itself is largely free from interstitially dissolved carbon.

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