

# INFLUENCE OF RETAINED AUSTENITE ON THE FRACTURE TOUGHNESS OF HIGH STRENGTH STEELS

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## ABSTRACT

A steel with nominal composition (wt%) Fe-0.4C-4Ni-2Si has been isothermally transformed in the bainitic temperature range and the microstructure characterised by transmission electron microscopy. The 'bainitic' structures contain interlath films of retained austenite (instead of cementite), and this leads to improved values of fracture toughness. It is suggested that the retained austenite films contribute to toughness by interrupting trans-packet crack growth, and that their ability to do this depends on their thermal and mechanical stability.

## KEYWORDS

Fracture toughness; high-strength steel; retained austenite; bainite.

## INTRODUCTION

Bainitic steels are not generally preferred for use in high strength steel applications in which high toughness is a primary requisite, in part because of the possibly embrittling nature of the bainitic cementite. However, it has recently been shown that the correct balance of alloy composition and heat treatment can produce 'bainitic' structures in which the formation of cementite is prevented by the partial stabilisation of the austenite (Matas and Hehemann, 1961; Oblak and Hehemann, 1967; Bhadeshia and Edmonds, 1979; Sandvik, 1982). This offers the possibility of producing tough bainitic steels using lean alloy compositions and simple heat treatments (Sandvik and Nevalainen, 1981; Bhadeshia and Edmonds, 1983a, 1983b).

There is still speculation, however, as to whether the toughness of such 'bainitic' steels can be attributed to the absence of cementite or to the presence of retained austenite (Bhadeshia and Edmonds, 1983a). It is the purpose of this paper to attempt to clarify the role of retained austenite in determining the resultant mechanical properties, particularly fracture toughness.

## EXPERIMENTAL PROCEDURE

The steel used was kindly supplied by RARDE, Fort Halstead, and the chemical composition is given in Table 1.

TABLE 1 Chemical Composition (wt%)

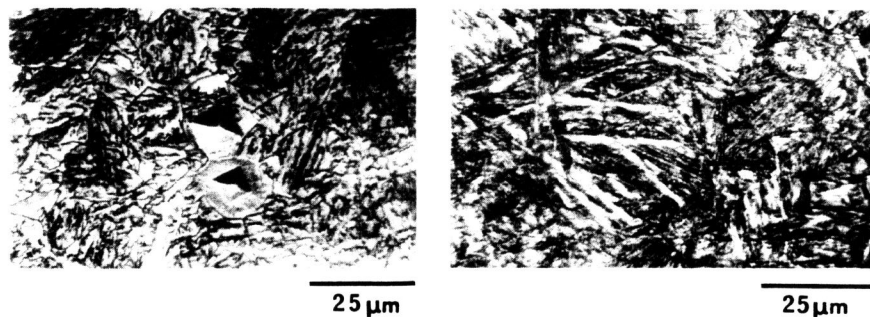
C	Ni	Si	S	P	O <sub>2</sub>	N <sub>2</sub>	Fe
0.40	4.15	2.01	0.0022	0.0016	0.0008	0.0003	balance

Compact-tension specimens were used to measure the fracture toughness in accordance with BS5447. A chevron notch was used to obtain a straight pre-crack in 25mm thick specimens. The specimens were pre-cracked in a fatigue testing machine and tested in a 100kN Instron machine at a crosshead speed of 1mm/min. Uniaxial tensile properties were measured using Hounsfield-type specimens 20mm in gauge length and 3mm in gauge diameter. The specimens were austenitised in a fluidised bath for 15min at 950°C. A Berkatekt coating was used to protect the specimens against decarburisation. Isothermal transformations were carried out by quenching into a molten salt pot containing stirred chemically-neutral salt, for times at least sufficient to complete the bainite reaction ( $\sim$  60min. for tensile specimens and  $\sim$  90min. for fracture toughness specimens). After isothermal transformation the specimens were water quenched.

Fracture surfaces were examined in a JEOL 35X scanning electron microscope. The retained austenite content was determined using a Philips X-ray diffractometer and Fe-filtered CoK $\alpha$  radiation and calculated from the integrated intensities taking into account the effects of texture (Dickson, 1969). After cutting and grinding, thin foils for transmission electron microscopy were thinned in a twin-jet electropolishing apparatus using 30-35V and a standard chromium trioxide-acetic acid solution (75g CrO<sub>3</sub> + 400ml CH<sub>3</sub>COOH + 20ml H<sub>2</sub>O) at room temperature.

## RESULTS

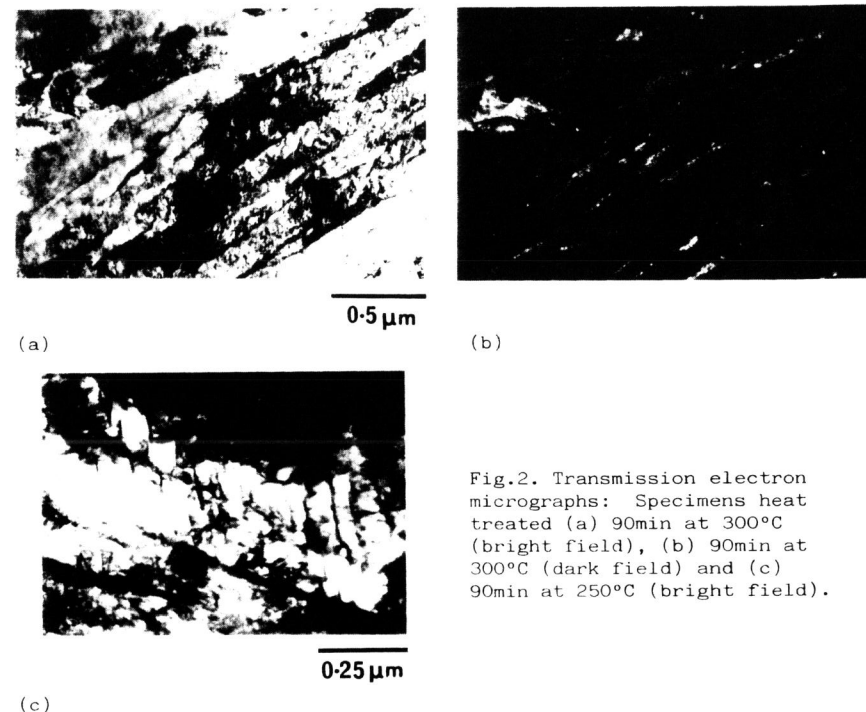
Optical micrographs of typical microstructures are shown in Fig.1. At the higher temperature (400°C), the transformation reaction did not result in



(a)

(b)

Fig.1. Optical micrographs: Specimens heat treated (a) 90min at 400°C, and (b) 90min at 300°C.



(a)

(b)

(c)

Fig.2. Transmission electron micrographs: Specimens heat treated (a) 90min at 300°C (bright field), (b) 90min at 300°C (dark field) and (c) 90min at 250°C (bright field).

complete decomposition, consistent with the incomplete reaction phenomenon exhibited by bainite (Hehemann, 1970), and large areas of blocky residual austenite were found in the structure. Transmission electron microscopy showed that some of the blocky residual austenite had transformed to plate martensite on water quenching.

Figure 2(a) shows a transmission electron micrograph typical of the transformation microstructure after heat treatment at 300°C. The microstructure consists of carbide-free bainitic-ferrite laths and interlath films of retained austenite, illustrated in dark-field in Fig.2(b). This structure can be considered as 'upper bainite', but in which the precipitation of interlath cementite has been prevented by the partial stabilisation of the austenite. According to a strict microstructural definition such carbide-free structures should not be classified as bainite. However, it is now recognised (Christian and Edmonds, 1983) that entirely ferritic structures can form in a way virtually identical to the formation of classical bainite in terms of morphology, kinetics and apparent growth mechanism. Furthermore, in the case of silicon steels, decomposition of the interlath films of retained austenite to carbide can be induced by further heat treatment which results in the more recognisable classical structure (Bhadeshia and Edmonds, 1979).

At 300°C thin films of retained austenite were the predominant austenite morphology, only a few isolated pools of blocky austenite being observed. The microstructure changes from upper to lower bainite as the transformation temperature decreases to 250°C. The 'lower bainite' structure is similar to

that after heat treatment at 300°C, except that precipitation of epsilon carbide has occurred within the ferrite plates (Fig.2(c)).

The mechanical properties and retained austenite contents are summarised in Table 2.

TABLE 2 Mechanical Properties and Retained Austenite Content

Transformation Temperature °C	YS MPa	UTS MPa	A %	Z %	$K_{Ic} \frac{1}{2}$ MPam <sup>3/2</sup>	$\gamma_R$ %
400	940	1172	17.9	46.4	47.0	20.6
300	1441	1648	12.0	66.4	95.0	8.7
250	1455	1834	11.2	50.8	81.2	12.1

YS = yield strength, UTS = ultimate tensile strength, A = elongation, Z = reduction in area,  $\gamma_R$  = retained austenite content.  $K_{Ic}$  values satisfy the validity criteria laid down in BS5447 (B = 25mm).

There is an increase in strength with decreasing transformation temperature. The reduction in area and fracture toughness are increased as the transformation temperature decreases from 400° to 300°C, but decrease again at 250°C. The elongation decreases with decreasing transformation temperature.

The fractographs illustrated in Fig.3 show that the final fracture mode

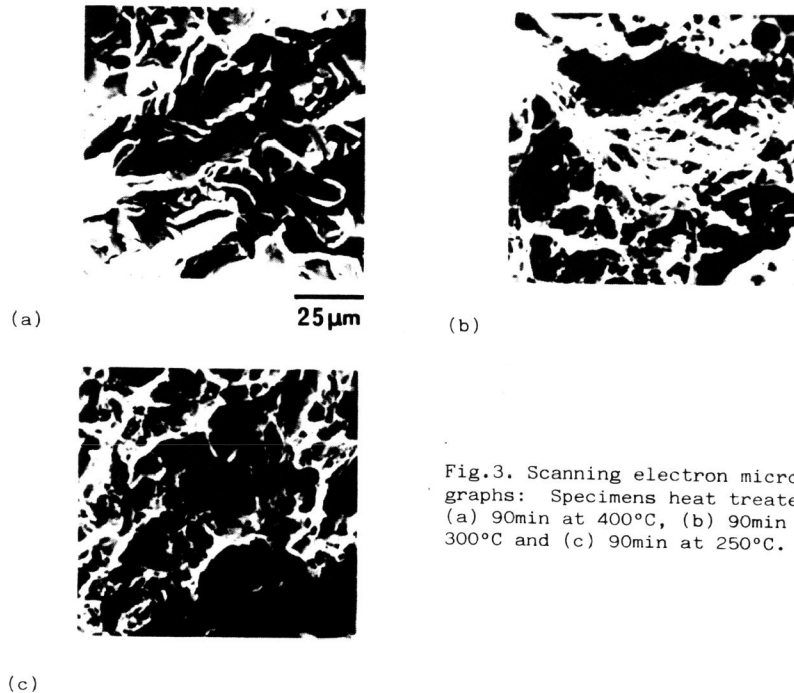


Fig.3. Scanning electron micrographs: Specimens heat treated (a) 90min at 400°C, (b) 90min at 300°C and (c) 90min at 250°C.

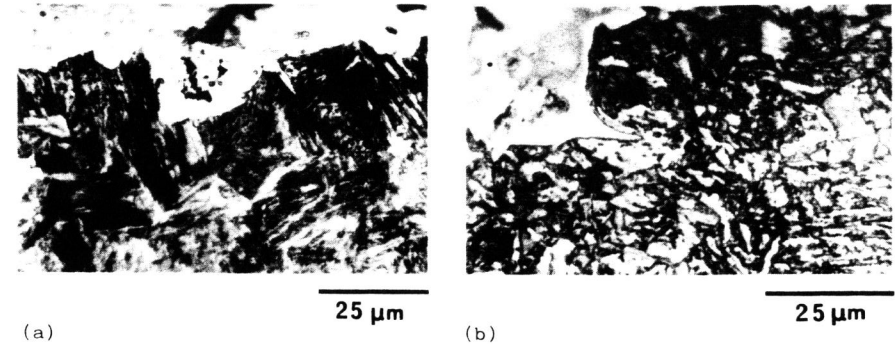


Fig.4. Optical micrographs: Specimens heat treated (a) 90min at 300°C and (b) 90min at 400°C.

changes from transgranular cleavage at 400°C to a mixture of dimple and cleavage fracture with decreasing transformation temperature. Examination of fracture surfaces by X-ray diffractometry did not reveal any retained austenite at the fracture surfaces, although the error involved in examining rough and plastically deformed fracture surfaces is expected to be large. Optical micrographs of sectioned nickel-plated fracture surfaces are shown in Fig.4.

#### DISCUSSION

Table 2 shows that fracture toughness of the upper bainite structure increases with decreasing transformation temperature. The drop in fracture toughness for the lower bainite structure formed at 250°C, as compared to the fracture toughness of upper bainite formed at 300°C, is probably related to the presence of the intralath carbides in the lower bainite, this being the most obvious distinction between the two otherwise similar microstructures. This demonstrates, contrary to normal expectations (Irving and Pickering, 1963; Pickering, 1967; Kalish, Cohen and Kulin, 1970; Thomas, 1973), that upper bainite can be tougher than lower bainite. This is a consequence of the substitution of retained austenite for the embrittling films of cementite normally present in upper bainite. Slatcher and Knott (1982) have studied tempered martensitic structures, and suggested that the initiation of voids at carbides is the important event which determines fracture toughness in high strength steels. The present results indirectly support this view, although a fine dispersion of carbides has been reported to contribute to toughness (Liu, 1969a, 1969b) suggesting that a critical carbide size may need to be exceeded before large reductions in fracture toughness are experienced.

The optimum fracture toughness is exhibited by the carbide-free upper bainite structure, in which retained austenite is homogeneously distributed along the lath boundaries (Fig.2(a)) rather than in blocky form. However, although the homogeneous distribution of interlath films of retained austenite seems to be beneficial to fracture toughness, the mechanism of toughening is not clear. It is known that very high stresses are encountered in the vicinity of the crack tip, and consequently retained austenite might be expected to transform to martensite under such conditions, as suggested by the X-ray results. Previous work (Syn, Fultz and Morris, 1978; Sastry, Khan and Wood, 1982) on martensitic alloys containing retained austenite has shown that retained austenite undergoes essentially complete decomposition to martensite in the plastic zone ahead of the crack. This localised, mechanically induced

martensitic transformation could be considered as an energy absorbing process which has led to a school of thought suggesting that localised martensitic transformation is responsible for the toughening. In the silicon-containing 'bainitic' steels, however, the interlath films of retained austenite have been shown to undergo substantial deformation without transformation to martensite (Bhadeshia and Edmonds, 1983a). Furthermore, the transformation of retained austenite in the upper bainitic structure would result in a brittle high-carbon plate martensite. It is, therefore, deduced that the beneficial effect of interlath films of retained austenite is not directly related to deformation-induced transformation.

Optical and electron microscopy showed that the optimum microstructure consists of packets of bainite which are made up of groups of similarly orientated sheaves consisting of parallel dislocated ferrite laths and thin films of interlath retained austenite. These packets are typically 10-15µm in size whereas the individual laths (in section) are  $\sim$  0.3µm in width. Examination of sectioned nickel-plated fracture surfaces suggested that considerable deviations of crack propagation path can occur within the bainitic packet, as illustrated in Fig.4(a). This is in contrast to conventional bainites in which the main direction of the crack propagation seems to be maintained across the packet (Naylor and Krahe, 1975; Lonsdale and Flewitt, 1978). It appears, therefore, that the role of interlath retained austenite is to interrupt the trans-packet crack growth, and therefore to refine the effective fracture grain size.

The blocky-type austenite is thought to be much less beneficial to fracture toughness, due to both thermal and mechanical instability, resulting in the formation of embrittling plate martensite and the cleavage fracture behaviour observed (Fig.3(a)). However, Fig.4(b) indicates that the blocky austenite is also able to stop (secondary) cleavage crack growth.

The above observations confirm that retained austenite can have a beneficial effect on fracture toughness, provided that it is thermally and mechanically stable, and therefore able to deform without decomposition to martensite, in order to restrict or blunt propagating cracks.

#### CONCLUSIONS

(i) Good fracture toughness obtained in a high-strength bainitic steel is attributed to the presence of thin films of thermally and mechanically stable interlath retained austenite. The role of the retained austenite is to refine the effective fracture grain size and to blunt the propagating crack.

(ii) Thermally and mechanically unstable blocky austenite and intralath carbides are detrimental to fracture toughness.

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