

The Fracture of Ausformed Steels

A. R. COX, P. JONES

Royal Armament Research and Development Establishment, Sevenoaks, Kent, UK

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The influence of ausforming on the methods of tensile and impact fracture in a low alloy steel, has been studied.

The ausforming produced austenite grains elongated parallel to the rolling direction and on subsequent quenching and tempering between 400 and 600° C, extensive networks of cementite particles formed along their boundaries. In this condition tensile failure occurred by the formation of longitudinal cracks along the prior austenite boundaries and these cracks joined together by transverse ductile tearing. On tempering above 600° C the boundary carbides coalesced and the matrix softened, and in this condition the steel possessed considerable resistance to transverse crack propagation.

1. Introduction

In recent years a number of thermal mechanical treatments have been developed to improve the properties of steels, and ausforming is one such treatment in which the steel is quenched from the austenitising temperature to 500 to 600° C, and deformed whilst still austenitic, before quenching to martensite. Only certain steel compositions are responsive to this thermal mechanical treatment, and these contain carbide formers, as well as possessing transformation characteristics which prevent transformation to bainite during the deformation of the metastable austenite. Once ausformed, the steels possess higher strengths than their conventionally treated counterparts, and this is obtained without an accompanying loss of toughness [1].

The factors contributing to the increased strength have received the attention of a number of workers, but there is still some doubt as to whether the increased dislocation density or the finer carbide dispersion in the ausformed steels is responsible for the improvement [2-4]. In contrast, the mechanisms responsible for the improved toughness have received less attention [5], and since ausforming offers one method of obtaining steels with greater resistance to crack propagation [6], the fracture processes in a series of ausformed low alloy steels were studied.

2. Materials and Treatment

The compositions of the two vacuum induction

melted steels used in the investigation are given in table I.

TABLE I

Cast	Composition of the steels wt %						
	C	Si	Mn	Mo	V	Cr	W
A	0.22	1.5	0.50	0.50	0.75	2.7	0.50
B	0.24	0.45	2.0	0.50	0.75	2.7	0.50

The two steels were austenitised for 1½ h at 1050° C, quenched into salt at 580° C, and after holding at this temperature for 5 min, were rolled to give a 69% reduction in area. The steels were then oil-quenched, stress-relieved at 300° C for 1 h and tempered for 1 h between 400 and 700° C. Test pieces were machined from the ausrolled 1.6 cm diameter bar, with the tensile and Charpy long directions parallel to the rolling direction [6].

3. Results

3.1. Metallography

Examination of the two steels in the optical microscope showed the structures to consist of light and dark bands 10 to 100 μm wide. These bands were the result of interdendritic segregation during the solidification of the original ingot, and electron probe microanalysis indicated that the chromium, manganese and vanadium contents of the dark bands were 25% greater than those of the light bands. As a result of these

compositional variations, the light bands were approximately 30 Knoop hardness numbers harder than the dark bands after tempering at 300° C, due to the presence of undissolved alloy carbides in the latter, whilst at the higher tempering temperatures the dark bands contained finer dispersions of alloy carbides and consequently were harder than the light bands.

A detailed study of the metallography using carbon extraction replicas showed that in the steels tempered at 600° C and above there were large differences in the distribution of carbide particles (fig. 1). In the light bands the martensite grain shapes were retained and there were few

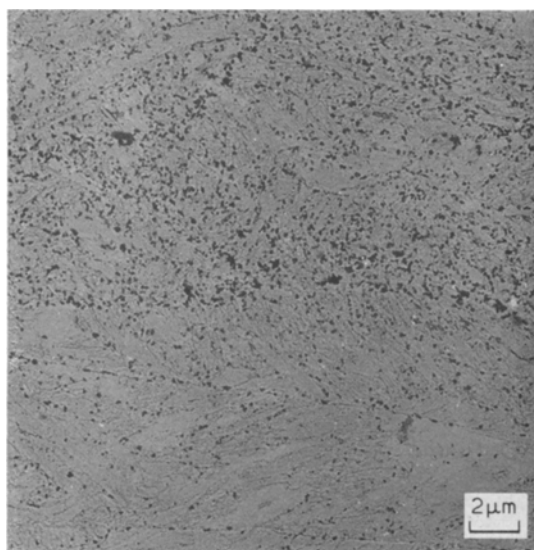


Figure 1 Steel A; ausformed and tempered 1 h at 700° C (extraction replica).

carbide particles, whereas in the dark bands the martensite contained numerous carbide particles. In all regions the elongated prior austenite boundaries were evident and those delineating the boundaries between the bands often contained long networks of carbides.

In the lightly tempered steels, although the banding was seen under the optical microscope it was not visible in the replicas. These showed the martensitic structures in the steels tempered at 300° C, contained Widmanstätten carbides, and after tempering at 450° C revealed the carbide particles to be at the martensite and prior austenite boundaries.

3.2. Tensile Fractures

In the tensile tests the pieces tempered between 400 and 600° C, delaminated after necking and

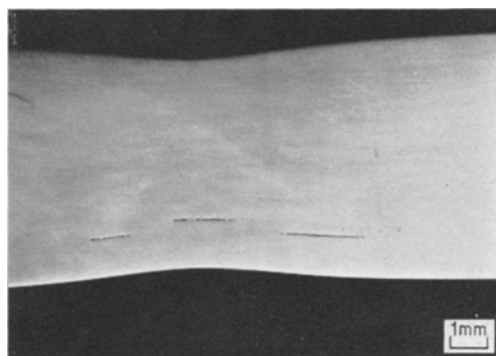


Figure 2 Steel B; ausformed and tempered 1 h at 300° C (optical micrograph).



Figure 3 Steel B; ausformed and tempered 1 h at 500° C (optical micrograph).

then fractured outside the necked portion, whereas those pieces tempered outside this temperature range broke conventionally in the necked region. During the necking, longitudinal cracks formed, and sections taken from test pieces, necked but not fractured, showed that the longitudinal cracks formed independently of transverse cracks (fig. 2). In the steels tempered outside the range 400 to 600° C, the longitudinal cracks were stable and did not propagate, whereas when tempered within this range some of the cracks were unstable and propagated 10 to 30 mm along the length of the test piece. Final transverse rupture then occurred as the longitudinal cracks joined together by transverse ductile tears (fig. 3) to give the serrated fracture topography seen in the Stereoscan (fig. 4). When

delamination occurred the surface of separation was flat and revealed the boundaries of the elongated prior austenite (fig. 5).

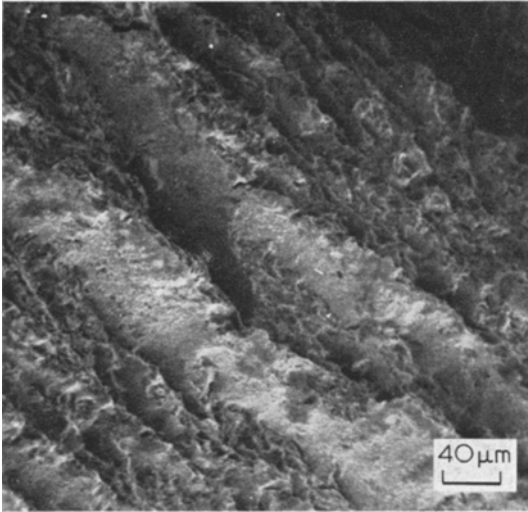


Figure 4 Steel A; ausformed and tempered 1 h at 550° C (scanning electron fractograph).

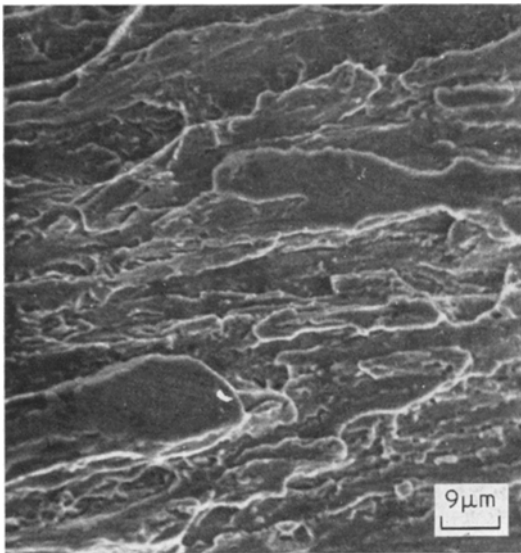


Figure 5 Steel A; ausformed and tempered 1 h at 550° C. Longitudinal crack surface (scanning electron fractograph).

3.3. Fracture of the Charpy Impact Test Pieces

On tempering above 400° C the toughness of the steels decreased, and minimum impact values at room temperature were obtained after tempering at 550° C. Further tempering then

increased the toughness, and values of the order of 90 ft lbs were achieved after tempering at 700° C (fig. 6). As the toughness increased, cracks propagating normal to the surface of the Charpy test piece began to show extensive longitudinal growth, giving a serrated fracture topography, similar to that observed in the tensile fractures.

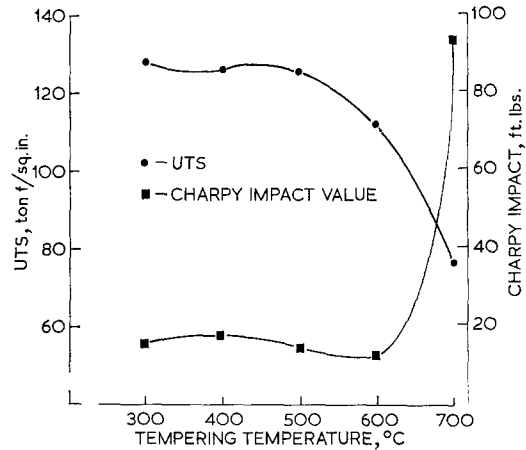


Figure 6 Variation of the UTS and Charpy impact values with tempering temperature for steel A.

Considerable plastic deformation accompanied fracture in the steels tempered at 700° C, voids formed along the bands ahead of the advancing cracks (fig. 7), and the steel ruptured along the bands by intergranular separation and ductile tearing; the latter being the predominant mechanism. These longitudinal cracks increased the path length of the propagating transverse cracks and also served to arrest propagating cracks by deflecting them into the longitudinal direction.

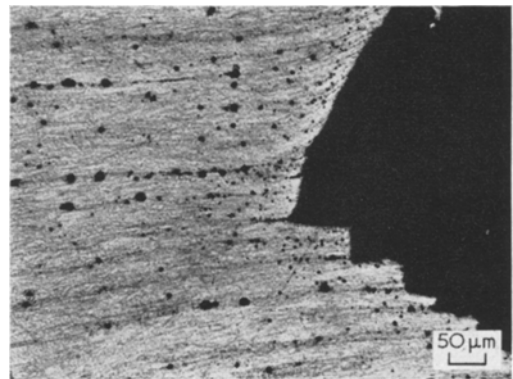


Figure 7 Steel A; ausformed and tempered 1 h at 700° C (optical micrograph).

4. Discussion

In the results of the tensile tests it was shown that a feature of the rupture process was the ease with which cracks initiated and propagated along the bands. This was noted especially in the steels tempered between 400 and 600° C, when considerable grain-boundary separation occurred. Previous work on the tempering of steels has shown that during tempering in this range the carbide particles in the martensite are dissolving and being replaced by films of carbides along the boundaries [7]; above 600° C these films coalesce and globular alloy carbides are formed [8]. The present steels conformed with this general pattern of decomposition, the principal difference being that the carbide precipitation along the austenite boundaries produced lines of carbide particles throughout the structure. These carbide particle networks were most extensive in the solute enriched dark bands.

Low [9] divided intergranular fracture into two types, firstly when a brittle phase is present at the grain-boundary, and secondly when impurity atoms segregate to the boundary. Of these two mechanisms, the former was responsible for the intergranular rupture in the present steels, separation being facilitated by the carbide particles present along the prior austenite boundaries. The embrittling effect of the carbide particles was exaggerated by the ausforming, for the flat elongated austenite grains produced during working facilitated the growth of extensive longitudinal cracks. Once the tempering temperature was increased above 600° C, the carbide films coalesced and the tendency for extensive delamination was reduced.

In a tensile test of a conventionally treated steel, fracture occurs by the formation and growth of voids in the necked region [9, 10]. As the stress increases the voids coalesce by internal necking and cracks propagate radially from the centre, the final rupture at the edge being by shear. Rupture by this mechanism produces the normal cup and cone fracture and explains the equiaxed and elongated dimples present in the centre and outer annulus respectively of a ruptured tensile test piece.

In the ausformed steels the first stage in fracture after necking is the growth of longitudinal cracks by intergranular separation along the prior austenite boundaries. Subsequent propagation of these cracks depends upon the nature of the carbide particle networks: the more continuous the networks the greater the crack growth,

with final rupture occurring by transverse ductile tears joining the longitudinal cracks together.

To account for longitudinal cracks in fibrous materials Cook and Gordon [11] discussed the stress distribution at the end of a transverse crack, with the applied load normal to the crack length. Under these conditions a tensile stress exists across the section and attains a maximum about one crack radius away from the crack tip. As a result of the tensile stress across the section, cracks can form by the separation of weak interfaces normal to the propagating crack, and when the transverse and longitudinal cracks join together the stress concentration of the propagating transverse crack is reduced. The authors propose that by this mechanism propagating cracks could be arrested and a tough material obtained. In the present investigation the ausformed steels were banded, and in this respect similar to fibrous materials, but in none of the tensile pieces was there evidence of transverse cracks forming prior to the longitudinal cracks. However, Cook and Gordon's crack blunting mechanism is probably responsible for the increased toughness of the steels tempered at 700° C, for considerable plastic deformation occurred ahead of the propagating crack and this produced separation along the bands normal to the crack growth.

The formula for the fracture strength (σ) of a brittle material modified for ductile fracture by including an additional term (γ_p) for the plastic deformation at the tip of the crack [12], is as follows:

$$\sigma = \left[(\gamma + \gamma_p) \frac{E}{C} \right]^{\frac{1}{2}},$$

where E is Young's modulus, γ is the surface energy and C is the crack length. When a steel ruptures by general intergranular separation the associated fracture energy is small, since there is little accompanying plastic deformation. In ausformed steels the intergranular separation only represents a small percentage of the total fracture process as it is restricted to the separation of the band interfaces and the high fracture energies result from the plastic deformation necessary to re-initiate the transverse ductile cracks which have been arrested at the separated band interfaces.

Grain-boundary separation was most extensive in the tensile test pieces tempered between 400 and 600° C, and not until the steels were

tempered above 600° C were the high fracture energies achieved. For Cook and Gordon's model to apply, the strength of the boundary needs to be less than a fifth of the molecular cohesion of the matrix. However if this were the only factor the ease of boundary separation in the tempering range 400 to 600° C would also be expected to result in large fracture energies. Since this was not found, it is considered that crack initiation after arrest at the interface is an important factor, and not until the alloy carbides have precipitated and coalesced and the matrix softened, does crack initiation in the matrix become difficult.

It is concluded that the ability of banded ausformed steels to arrest cracks depends upon careful control of the matrix properties and the distribution of boundary carbides. Only by controlling these can the tendency of the steels to delaminate be used to improve their toughness.

5. Conclusions

(i) When ausformed steels are tempered, carbide particles form along the elongated austenite boundaries and in the solute enriched bands produce extensive lengths.

(ii) If ausformed steels are tempered in the range 400 to 600° C films of carbide particles form along the austenite boundaries and facilitate intergranular separation. When tempered above 600° C the carbides coalesce and under impact loads these particles nucleate voids which arrest the growth of transverse cracks.

(iii) In banded ausformed steels the first stage in the tensile rupture process is the nucleation and growth of longitudinal cracks by intergranular separation, and final rupture occurs when the longitudinal cracks join together by ductile tearing.

Acknowledgement

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