Fracture of martensite with particular reference to ausformed martensite

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SYNOPSIS

The fracture characteristics of a 12%Cr steel (En 56C) in the aus-formed and conventional conditions have been studied. Mechanical test data have been correlated with details of fracture surfaces and microstructure of the martensite.

Ausforming (and subsequent tempering) brought about a beneficial effect on the strength and toughness of the steel, and at a given strength level the ausformed material was considerably tougher than the conventionally treated steel.

An electron microscope study of fracture surfaces and fracture profiles revealed two basic modes of fracture, quasi-cleavage and ductile rupture. In general, ausforming brought about an increase in the proportion of ductile fracture which could be correlated (in the 400°–600°C tempering range) with an increase in the absorbed impact energy.

It has been shown that the improvement in toughness brought about by the ausforming process was derived from the precipitation of a fine, well dispersed chromium carbide phase (with an accompanying effect on the dislocation configuration) during the working operation. 2617

INTRODUCTION

IT IS well established that the strength of certain martensitic steels is considerably improved when the material is deformed in the metastable austenite condition before transformation to martensite.[1-5] This treatment (ausforming) has also been used to increase the ductility of certain high-strength steels at a given strength level. Previous work[4,6] at BISRA demonstrated the ausforming response of several alloy steels (En56C, En30B, 6%Mn–3%Cr). The present investigation was carried out to examine the effect of ausforming on the fracture resistance of these steels, and to study the fracture characteristics of the martensite.

The electron microscope can be used with great advantage in the examination of fracture surfaces, since the instrument combines a large depth of field with high resolution. In the present work an attempt is made to correlate details obtained by electron fractography with fracture resistance and microstructure.

A systematic investigation involving mechanical testing, electron fractography, and thin-foil work was carried out on each of the steels mentioned above. In general, an ausforming treatment brought about an improvement in the fracture resistance of all the materials investigated. This improvement was enhanced (in varying degrees) by a subsequent tempering treatment. At all stages of the examination the ausformed martensite was compared and contrasted with a conventionally treated martensite. For the sake of clarity only the results obtained with a 13%Cr steel (En56C) will be dealt with in this paper.

TECHNIQUES

The material (En56C) was supplied from commercial stock in the form of 9/8in dia. bar. The analysis (wt-%) is given in Table I. For the ausforming treatment the 9/8in bar was austenitized at 1 000°C for 1 h and then quenched into a lead bath maintained at 500°C (the optimum ausforming temperature[4]). The TTT diagram for this steel shows an extended austenite 'bay' at this temperature. The bar was ausformed on a draw bench from 9/8in to 5/8in dia. in 1/16in passes. The material was returned to the lead bath (for 3 min) after each pass to keep the temperature of deformation as constant as possible. The bar was air cooled after the last pass. The total amount of ausdeformation was ⁓70%. Samples of conventionally treated martensite were obtained by hot rolling some of the 9/8in bar down to 5/8in diameter. The material was then austenitized at 980°C for 1 h and air cooled.

After machining, the specimens (Charpy and tensile) were tempered for 1 h periods in the range 200°–600°C. The tempering treatments were carried out under an atmosphere of argon. The specimens were then tested at room temperature. The Charpy specimens were broken in a machine delivering a

TABLE I   Analysis, wt-%, of En56C bar

Cr	Si	Mn	Ni	Mo	V	C	Fe
12·65	0·34	0·27	0·06	0·017	0·01	0·23	Balance

striking force of 216 ft lb. Immediately after testing, self-shadowed carbon replicas were prepared from the fracture surface of one half of the broken testpiece. The topography of the fracture face gave a 'shadowing' effect which revealed the relative heights and depths of the features on the fracture surface. To study the fracture profile and its relation to the underlying microstructure, the fracture face on the other half of the test piece was plated with nickel, and a longitudinal section taken through it. Carbon extraction replicas were taken from the Ni/specimen interface.

The broken Charpy testpieces were electropolished for optical microscopic examination and hardness tests. For detailed structural examination, thin foils were prepared from the Charpy test pieces. A mechanical-grinding, chemical-polishing technique[7] was used to obtain strip 0·006in in thickness. The final thinning was accomplished by electropolishing in a modified Bollmann apparatus,[8] using a chromic oxide–acetic acid electrolyte.[9] The full details of the thin foil preparation technique are given elsewhere.[10]

RESULTS
Mechanical tests
The mechanical test data for ausformed and conventionally treated specimens are given in Figure 1 (*A–D*). The data on tensile strength and elongation of ausformed En56C have been taken from an earlier paper,[4] but some of the tests were recently duplicated and the results agreed very well with those obtained in the previous work.

It can be seen that ausforming increased the strength (Figure 1*A* and 1*C*) of the steel without an accompanying drop in impact absorbed energy (Figure 1*D*). This effect was maintained on subsequent tempering in the 100°–400°C temperature range. Over this 'low-temperature' tempering range the tensile strength of both the conventionally treated and the ausformed steels decreased by about 10 tons/in² while the Charpy impact-absorbed energy remained substantially the same.

On tempering in the 400°–600°C temperature range a marked change in the mechanical properties was observed. The tensile strength fell drastically, but even after tempering at 600°C the ausformed material was stronger than the conventionally treated steel (73 tons/in² against 58 tons/in²). The impact energy of the ausformed steel increased rapidly and after tempering at 600°C was over 100 ft lb (compared to 22 ft lb for the conventionally treated sample). In other words, after tempering at 600°C the ausformed material was a little stronger and much tougher than the identically tempered conventionally treated material.

Electron fractography
It was felt that the difference in fracture toughness between the ausformed and the conventionally treated specimens warranted a detailed examination of the fracture faces. Replicas of fracture faces and fracture profiles were examined in the electron microscope.

Terminology
Depending on the heat treatment, the martensite exhibited ductile and/or brittle fracture characteristics. In the brittle case the extensive flat facets (which are typical of low energy fracture) cannot be correlated (in terms of size) with any structural
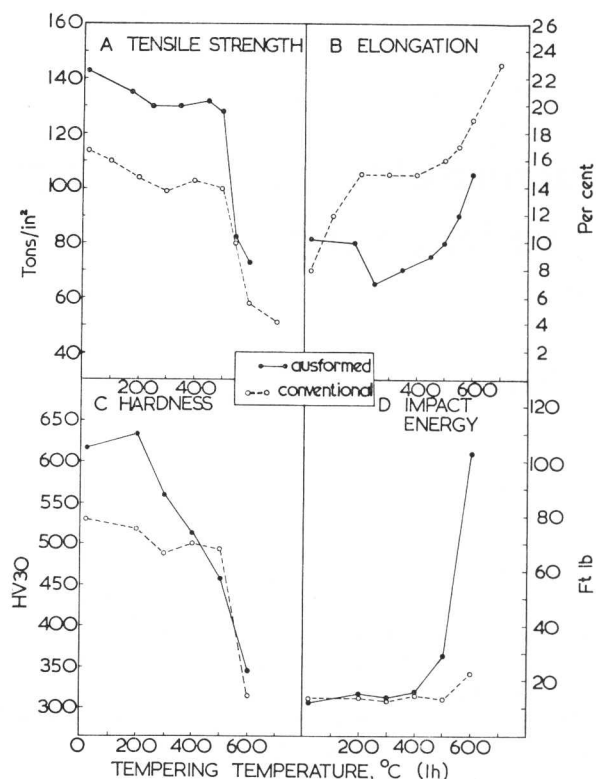
*A* tensile strength; *B* elongation; *C* hardness; *D* Charpy impact absorbed energy

1 *Mechanical test data on En56C (ausformed and conventionally treated)*

feature of the fine martensite units. It was obvious that these large facets, resembling cleaved surfaces, could not be true 'cleavage' fractures, and could not be attributed to failure along any particular lattice plane of the bct (martensite) or bcc (tempered martensite) lattices. Because of the absence of any definite relationship between the fracture facets and the lattice planes, the term 'quasi-cleavage' has been proposed[11] for this type of fracture in martensitic steels.

Ductile fracture grows by the coalescence of internal voids, and is characterized by the appearance of 'dimples' or 'cusps' on the fracture surface. These 'dimples' may be equiaxed (when the voids are spherical) or parabolic (when the voids are ellipsoidal) in appearance. Under the application of a non-uniform stress, such as that present during a Charpy impact test, the ductile fracture grows by the coalescence of voids at the tip of the advancing crack (*see* Figure 14). Such a mode of crack growth has been termed 'tearing'. The dimples produced during 'tearing' are parabolic (i.e. elongated) in shape and point towards the fracture origin (Figure 14). Both types of fracture referred to above are transgranular.
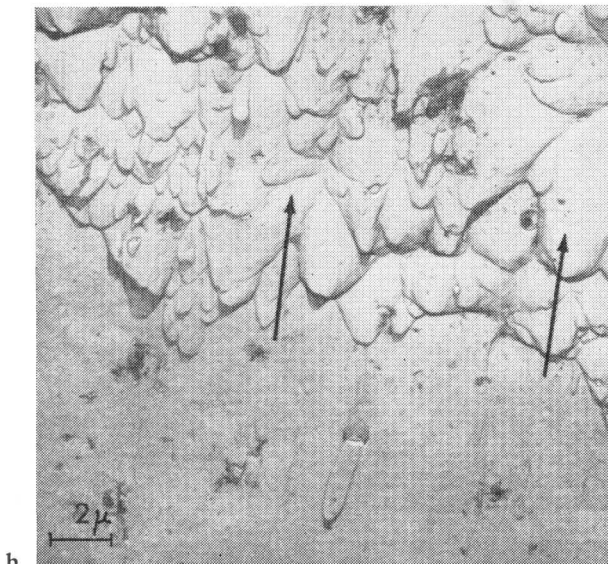
Description of fracture surfaces
The ausformed (Figure 2*a*) and conventionally treated (Figure 2*b*) samples in the as-quenched state showed similar fracture characteristics. The fracture surface was mainly composed of large facets of quasi-cleavage fracture with small areas of ductile fracture. The elongated dimples making up the ductile
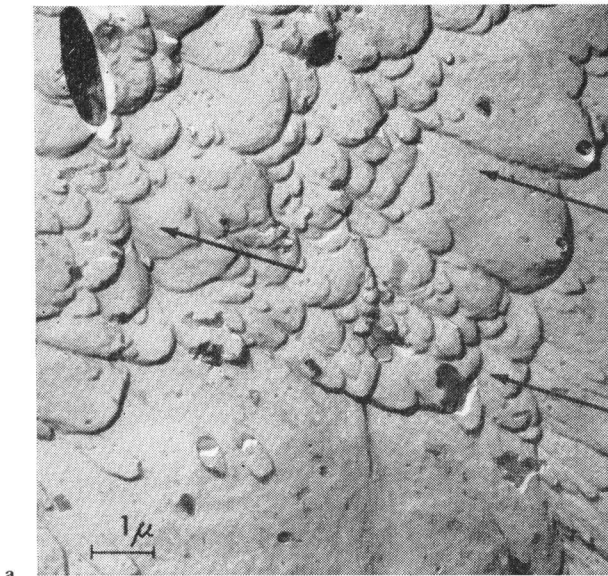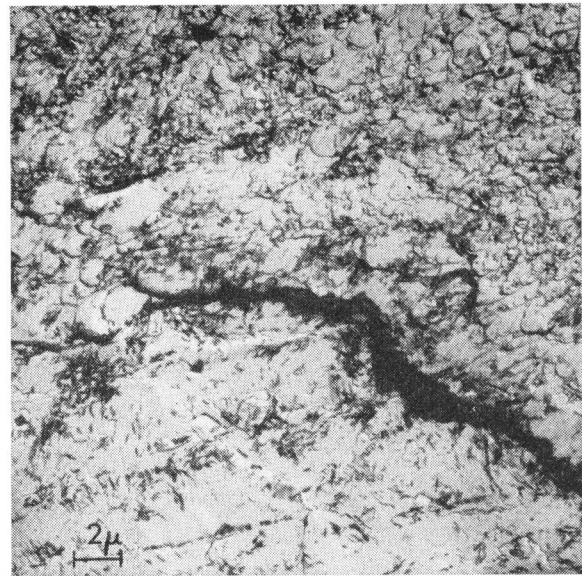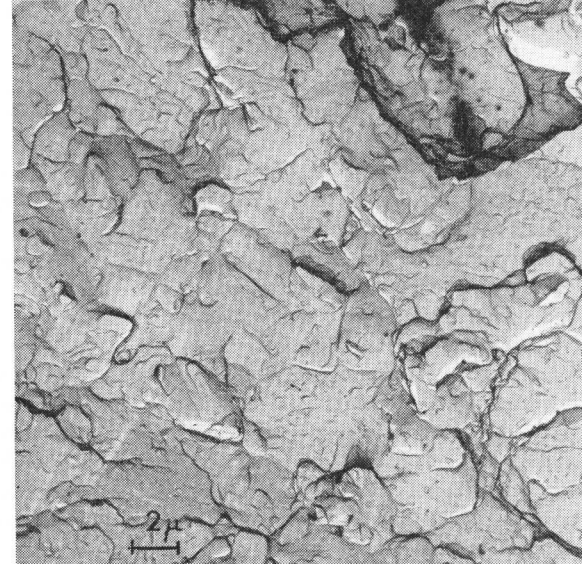
a ausformed – untempered (×8 000); **b** conventionally treated – untempered (×4 000)

2 *Electron fractographs; arrows indicate direction of crack propagation*

a ausformed – tempered for 1 hour at 400°C (×3 000); **b** conventionally treated – tempered for 1 hour at 400°C (×3 000)
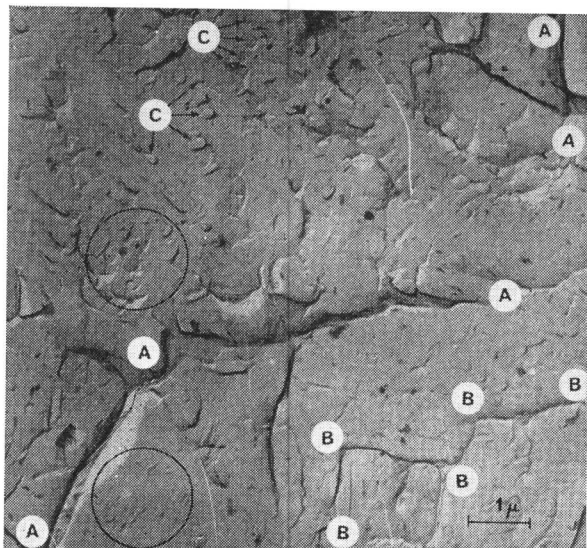
3 *Electron fractographs*

regions indicate the direction along which the fracture travelled (shown by the arrows). One difference between the two samples was that in the ductile regions the size of the fracture units (i.e. dimples) was larger in the conventional material than in the ausformed.

A tempering treatment at 200° and 300°C (for 1 h periods) did not bring about any substantial change in the details of the fracture surfaces. The ausformed samples showed more extensive dimpled areas, while the fracture of the conventional material was predominantly quasi-cleavage.

After tempering for 1 h at 400°C the fracture topography of the two samples was quite different. The ausformed specimen showed a 50 : 50 ductile/quasi-cleavage fracture surface, and a coarse precipitate of cementite was associated with each type of

fracture (Figure 3a). The conventionally heat treated specimen again showed a quasi-cleavage fracture and no precipitate was seen on the fracture surface. The flat transgranular fracture facets could be seen quite clearly on the Ni-plated fracture profiles (*see* Figure 9b).

The fractures observed after tempering at 500°C were very similar to those at 400°C, except that in the ausformed specimen there was no Fe_3C associated with the fracture surface. Figure 4 shows the quasi-cleavage fracture of the conventionally treated martensite at this stage of the tempering (1 h at 500°C). This figure shows all the surface characteristics that are attributed to quasi-cleavage.[12] The dark and heavy markings between points A – A are caused by 'tear ridges', which are formed by the linking up of submerged cracks with one another

4   *Conventionally treated – tempered for 1 hour at 500°C; characteristic topography of quasi-cleavage fracture surface; electron fractographs*   ×15 000

or with the advancing fracture front. The latter case is shown in section in Figure 13 at points marked with a 'T'. The lines between points B–B are 'steps' which are formed on the fracture face as the crack propagates across the grain ('S' in Figure 13). The small circular markings ('tongues') on the flat facets (at points marked C in Figure 4) are also a characteristic of quasi-cleavage. River markings can be seen quite clearly in the circled areas. The inference is that for these facets the fracture started within the facet itself rather than at the edges – where fracture would originate in true cleavage.

A tempering treatment of 1 h at 600°C resulted in the types of fracture shown in Figures 5 (*a* – *c*) and 6 (*a* – *b*). The ausformed samples showed an extremely ductile fracture with very small dimples, and a $(Cr, Fe)_7C_3$ precipitate was associated with the fracture (Figure 5*a*). The ductile dimples can be seen in section in the Ni-plated fracture profile (Figure 6*a*). The conventionally treated sample showed a 50 : 50 quasi-cleavage/ductile fracture (Figure 5*b* – *c*) and there was considerable precipitate in the 'grain' boundaries (i.e. the martensite 'bundle' boundaries). The coarse grain boundary precipitate can be seen in the fracture surface (Figure 5*b*) and in the fracture profile (Figure 6*b*). This specimen showed the only traces of intergranular fracture that were observed in this work. In the ductile areas (Figure 5*c*) the dimple size was much coarser than that observed in the corresponding ausformed sample (Figure 5*a*).

Structural examination

To explain the difference in toughness (and the accompanying differences in fracture topography) between the ausformed and conventionally treated martensites, it was necessary to carry out a thorough metallographic examination.

Optical examination showed that a coarse, widely dispersed precipitate was present in most of the specimens (Figure 7) and that the prior austenite grains in the ausformed specimens were elongated in the direction of working.
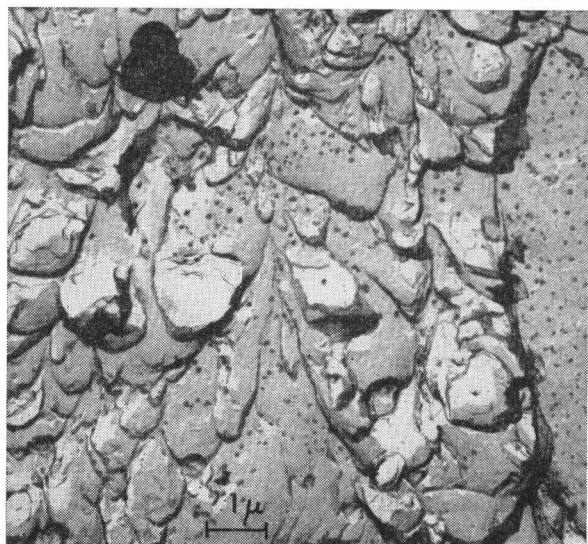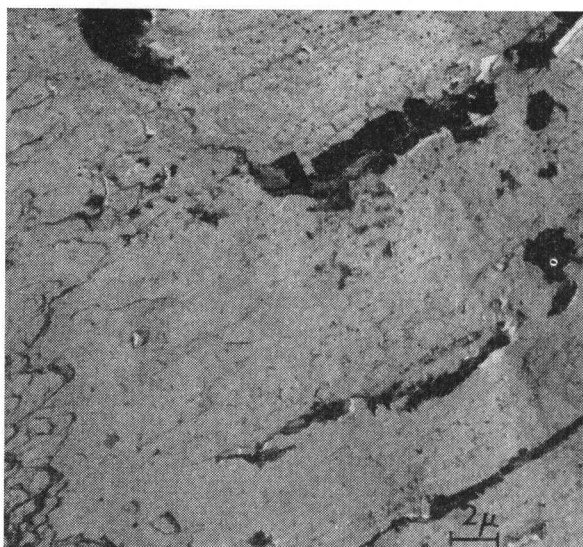
a ausformed – tempered for 1 hour at 600°C (×8 000); **b** conventionally treated – tempered for 1 hour at 600°C (×3 000); **c** conventionally treated – tempered for 1 hour at 600°C (×8 000)
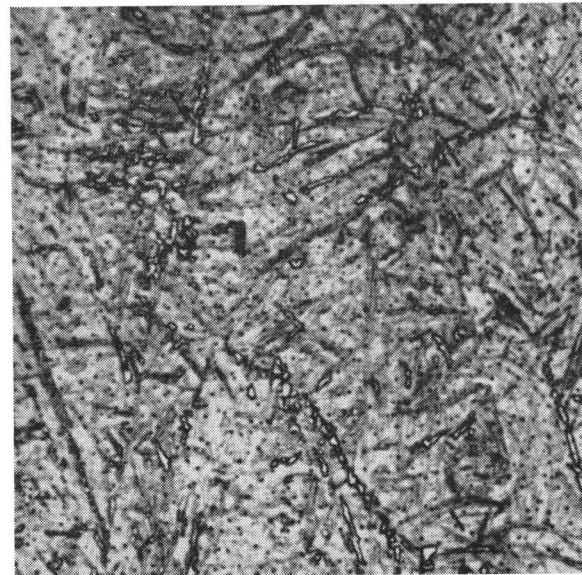
5   *Electron fractographs*

7   *Ausformed; as–quenched; optical micrograph*   ×800

a ausformed – tempered for 1 hour at 600°C (×6000); **b** conventionally treated – tempered for 1 hour at 600°C (×6000)
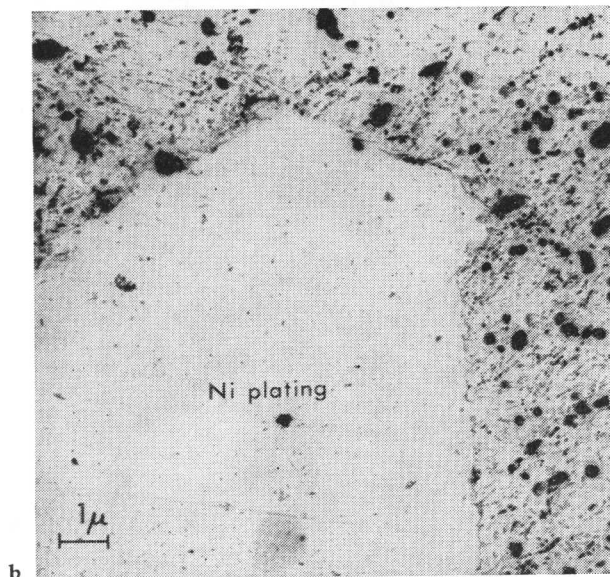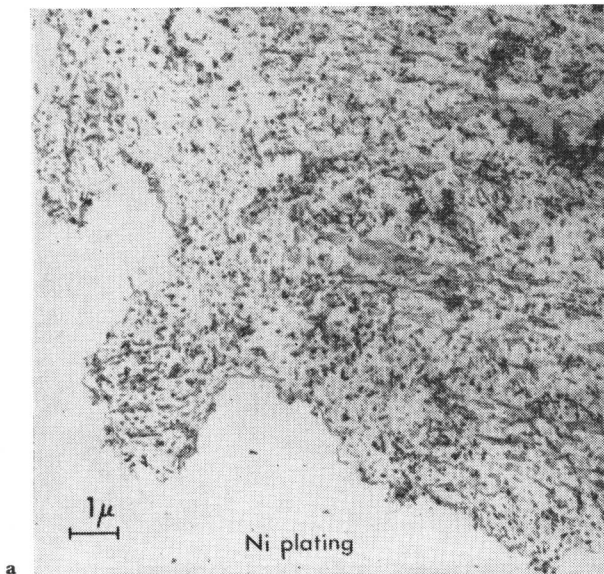
6   *Ni–plated fracture profile; extraction replicas*

The finer details of the martensitic structure (ausformed and conventionally treated) are shown in Figures 8–12. The figures show that each plate seen under the optical microscope is actually a 'bundle' of very fine lamellae, as has been well established for a steel containing ∼0·2 wt-%C.[14-16] The fine lamellae are referred to as martensite 'units'. In agreement with previous work[15,16] on low-carbon martensites, microtwins were not observed in the martensitic structure.

In the as-quenched condition the ausformed specimen showed a very high dislocation density (Figure 8a). In fact, in many areas the density was too high to resolve individual dislocations, and a 'mottled' contrast effect was observed. In comparison, the conventional material showed a lower (though still high) dislocation density (Figure 8b). The ausformed specimen showed clear evidence of a very fine precipitate in the martensitic structure (Figure 8a) but this precipitate did not give an identifiable diffraction pattern, nor was it possible to extract it on carbon replicas. Tempering in the range 200–300°C resulted in the precipitation of cementite in both types of martensite.

Figure 9 (*a*, *b*) shows the microstructure of the ausformed (9a) and conventionally treated (9b) martensites after tempering at 400°C. The distribution of cementite plates is more dense in the ausformed sample. Further structural details in the ausformed martensite tempered at 400°C can be seen in Figure 10. The large plates are cementite, and in the background there is another very fine precipitate. This second precipitate is presumably chromium carbide, but definite identification by electron diffraction was not possible because of the small volume fraction of the precipitate.

No cementite was detected in the ausformed and conventional martensites after tempering at 600°C. The carbon extraction replicas showed that the chromium carbide was more uniformly distributed in the ausformed sample (Figure 6a) than in the conventionally treated specimen (Figure 6b). In the latter the martensite unit boundaries were preferred sites for carbide precipitation. Also, the very coarse precipitates observed in the conventionally treated material (Figure 6b) appeared to be absent in the ausformed sample (Figure 6a). Thin-foil examination confirmed these observations. In the ausformed sample the carbide precipitate tended to be globular in form although a few elongated particles were seen along the martensite unit boundaries (Figure 11a and b). Once again, a very fine precipitate, generally associated with the dislocation arrays, was seen in the ausformed martensite (Figure 11 circled areas). Figure 11a and b shows the same specimen area in different tilt positions. The importance of tilting the foil to show up these very fine precipitates (the contrast of which is very sensitive to the reflecting conditions) is clearly demonstrated by this
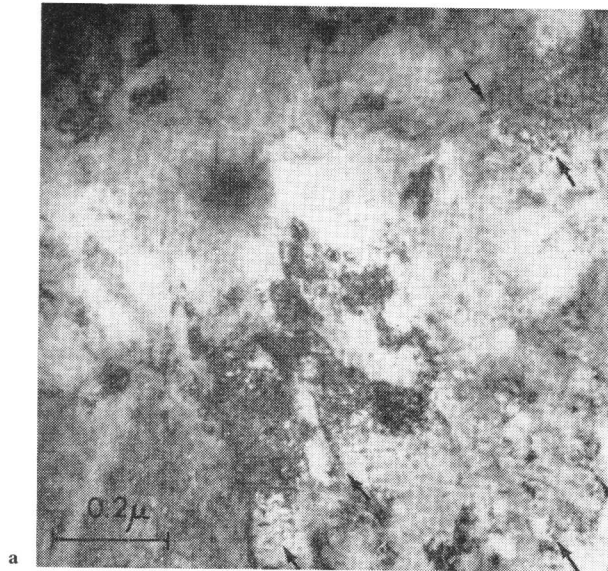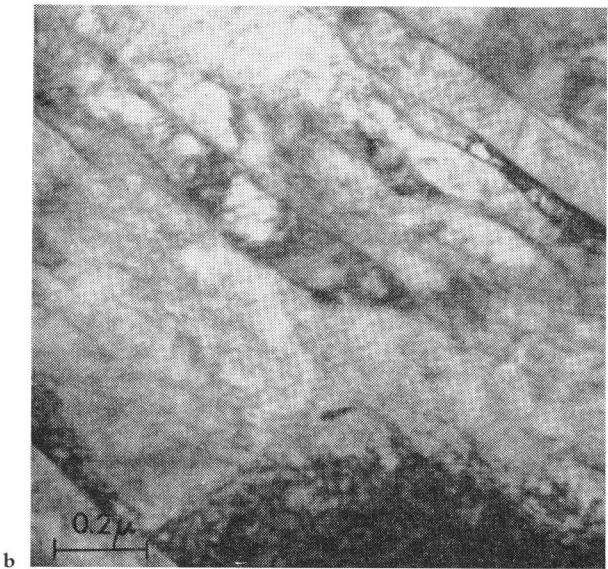
a ausformed; as-quenched; fine precipitate (arrowed) can be detected; thin foil ($\times 75\,000$); **b** conventionally treated; as-quenched; thin foil ($\times 60\,000$)
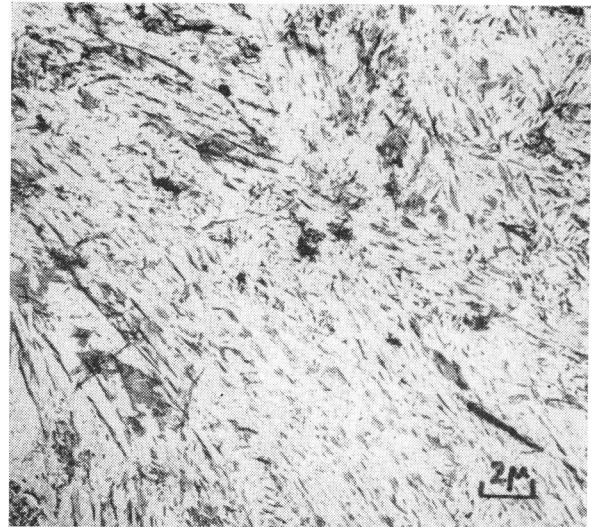
8   *Ausformed and conventionally treated martensite*

a ausformed – tempered for 1 hour at 400°C; carbon extraction replica ($\times 3\,000$); **b** conventionally treated – tempered for 1 hour at 400°C; carbon extraction replica ($\times 3\,000$)

9   *Ausformed and conventionally treated martensite*

pair of micrographs. These fine precipitates were not observed in the conventionally treated martensite (Figure 12*b*).

At this tempering temperature (600°C) the ausformed martensite showed the first signs of recovery (Figure 12*a*). The formation of dislocation networks in some of the martensite units indicated that dislocation climb had taken place at this temperature. The conventionally treated martensite also showed signs of recovery (Figure 12*b*).

An attempt was made to measure the sizes (length and width) of the martensite bundles and units in the ausformed and conventionally treated specimens. The average values are shown in Table II.

It must be pointed out that there was a very large range of sizes, and also that the width of the martensitic units and bundles could not be estimated with any degree of accuracy, since the angle at which they intersected the foil surface was not known.

DISCUSSION

The limitations of the Charpy impact test (due to design of specimen, notch acuity, etc.) in measuring the absolute toughness of a material were fully appreciated. However, this did not affect the main observation that had resulted from this work, which was that, at a given strength level, the ausformed martensite was superior in toughness to the conventionally treated martensite. This superiority was particularly marked on subsequent tempering in the range 400°–600°C. An attempt will be made to explain this difference, and to correlate the mechanical properties of the martensites with their microstructural features.

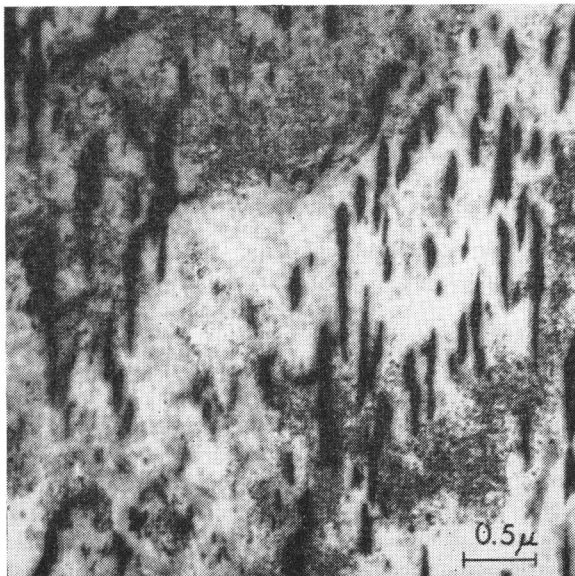**TABLE II Sizes of martensite bundles and units, average values**

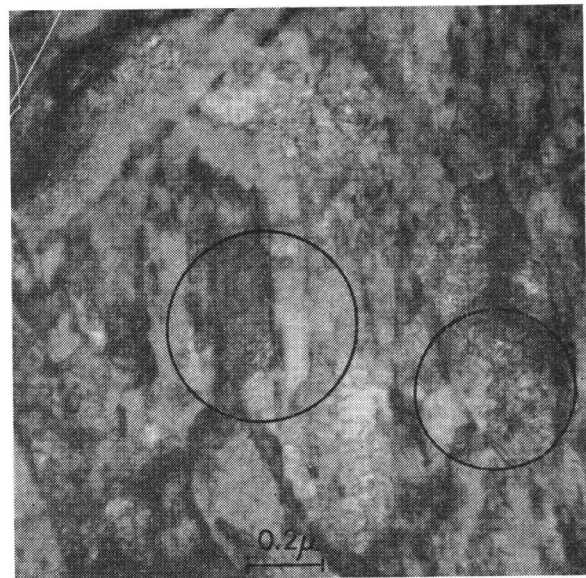| Martensite	Units		Bundles	
	L	W	L	W
Ausformed	1·3μm	0·1μm	5·5μm	12μm
Conventionally treated	4μm	0·4μm	—	—
L=length, W=width				

Fracture mechanisms

Two fracture modes have been observed in this work, 'quasi-cleavage' and ductile fracture. In the impact test the former is expected to absorb a lesser amount of energy than the latter. This correlation has been confirmed, the amount of ductile fracture (as seen by electron fractography) increasing with absorbed impact energy.

Under the application of a non-uniform stress (such as that which is applied during the Charpy tests) both quasi-cleavage and ductile fracture grow by the nucleation and growth of submerged cracks ahead of the advancing fracture front. The crack front advances in a stepwise manner as the voids in front of the fracture tip grow until coalescence takes place (Figures 13 and 14). The advance of a quasi-cleavage fracture crack through a martensitic structure is shown schematically in Figure 13. The fracture is transgranular with respect to the prior austenite grains and bears no definite orientation relationship to the martensite lattice. The quasi-cleavage fracture surface shows fractures characteristic of both true cleavage and plastic rupture. Firstly, extensive flat facets are seen on fracture surfaces and fracture profiles (sketched in Figure 13). The initiation of these facets is a feature of true crystalline cleavage. So also is the formation of 'steps' (points 'S' on Figure 13) during the growth of

a, b different tilt positions of the same area; thin foil (×50000)

11 *Ausformed; tempered for 1 hour at 600°C; very fine chromium carbide precipitates can be detected by tilting the foil*

10 *Ausformed; tempered for 1 hour at 400°C; the fine chromium carbide precipitate can be seen in between the cementite plates; thin foil* ×20000

the facet. Secondly, tear ridges are also observed on the fractographs (Figure 4) and are illustrated in Figure 13 at points marked 'T'. These are formed during the linking up of microcracks with the advancing fracture front. In this respect, quasi-cleavage resembles plastic rupture.

In ductile fracture the dimpled topography is due to the concave depressions formed by the growth and coalescence of spherical micro-voids with the advancing crack front (Figure 14). These micro-voids may be nucleated at any heterogeneity in the matrix (e.g. at the interface between precipitate and matrix). No correlation was found between the spacing of the precipitate particles and the size of the dimples. In most cases the diameter of the dimple was large enough to suggest that the
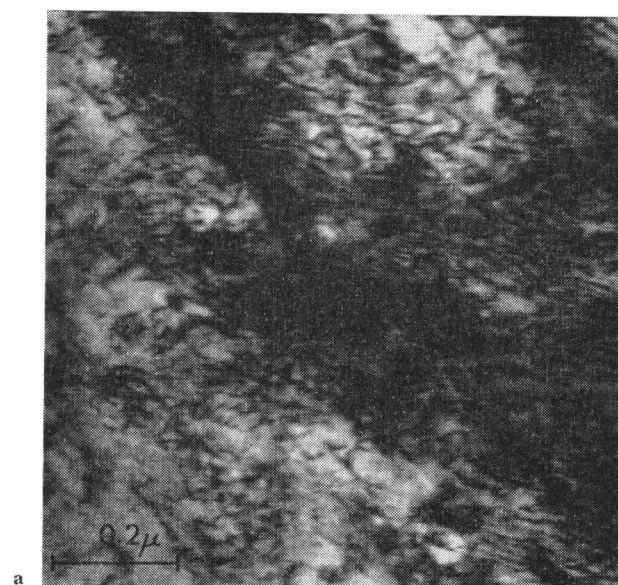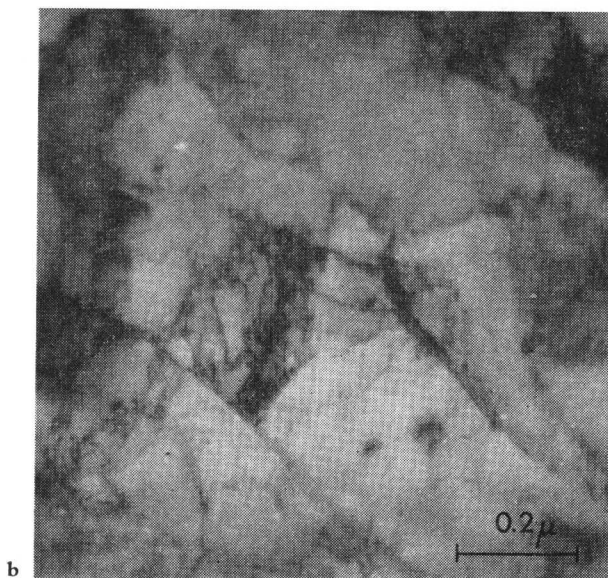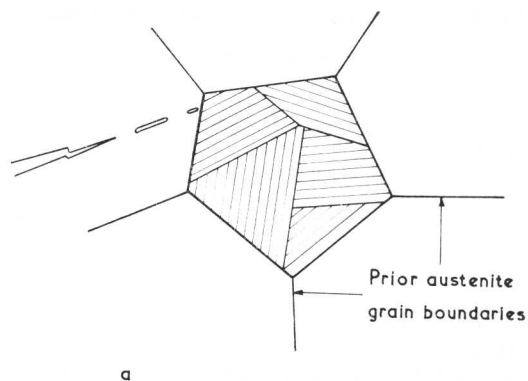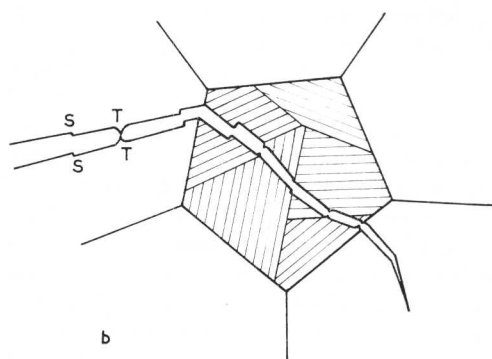
a ausformed – tempered for 1 hour at 600°C; thin foil (×80000);
b conventionally treated – tempered for 1 hour at 600°C; thin
foil (×80 000)

12   *Ausformed and conventionally treated martensite*

13   *Propagation of quasi-cleavage fracture through a martensitic structure (under non-uniform stress)*

micro-void had grown across the width of several martensite 'units' (as shown in Figure 14). The size to which a micro-void can grow depends partly on the work hardened state of the matrix, and so the number of voids required for the propagation of a fracture front will increase with an increase in the work hardened condition of the matrix. This accounts very well for the relative sizes of dimples seen in Figure 5a and c. As would be expected the dimples are finer in the ausformed sample, which, of course, is in a greater work hardened condition than the conventional martensite. The need to nucleate a larger number of micro-voids to propagate the fracture would also explain qualitatively the higher absorbed impact energy of the ausformed martensite relative to the conventional martensite.

Mechanical properties and microstructure

The coarse (globular) precipitates (Figure 7) seen in the microstructure of most of the samples were presumably alloy carbides which had not been taken into solution during the austenitizing treatment. These were too widely dispersed to have any significant effect on the strength and toughness of the material.

Untempered

The very finely dispersed chromium carbides detected in the as-quenched, ausformed sample had probably been precipitated from the metastable austenite during the thermo-mechanical treatment.[13] The fact that *all* ausformed samples (Figures 8a, 10, 11), but *none* of the conventionally treated martensites (Figures 8b, 12b) showed this fine precipitate, lends very strong support to this view. This precipitate (formed in the austenite) must have been 'inherited' by the martensite when the austenite → martensite transformation took place. The precipitate must also affect the dislocation distribution, and thin-foil examination of as-quenched, ausformed martensite did show that they had a higher dislocation density than the conventionally treated martensite. These fine precipitates (with the accompanying change in dislocation distribution) would account for the higher strength of the untempered ausformed martensite (∼143

Prior austenite grain boundary

Martensite "bundle" boundaries

a

b

c
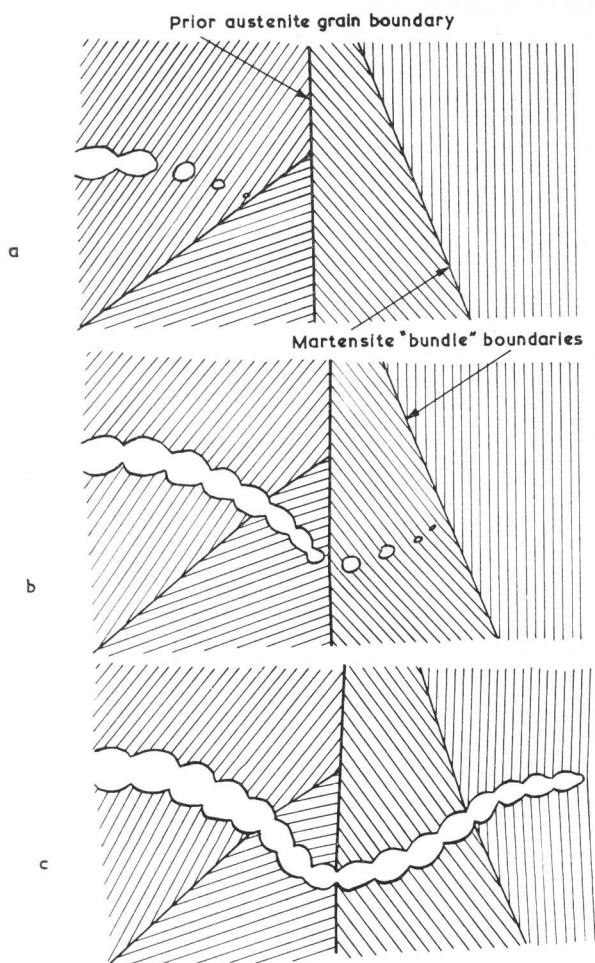
14 *Propagation of ductile fracture through a martensitic structure (under non-uniform stress)*

tons/in²) relative to that of the as–quenched conventional martensite (∼114 tons/in²). On the other hand these microstructural differences would not be expected to affect the toughness of the as-quenched material.

100°–400°C temper
During low-temperature tempering (100°–400°C), the fine chromium carbide precipitates were not affected. Precipitation of small plates of cementite did not alter the mechanical properties to any appreciable degree, except that the tensile strength of both types of martensite was slightly reduced. This drop could be explained by the lowering of the inherent strength of the martensite matrix as it became depleted in carbon.

400°–600°C temper
The most marked differences between the properties (and microstructure) of the ausformed and conventional martensites occurred after tempering in the range 400°–600°C. After tempering at 600°C there was no cementite in either of the martensites, but the dispersion of chromium carbide was much finer in

the ausformed sample than in the conventional sample. This difference in alloy carbide morphology and distribution can be explained on the basis that the ausformed martensite prior to tempering already had a fine and uniform dispersion of carbide in the matrix. These must grow during tempering, and, since there was such a high density the growth of each carbide particle was severely restricted (i.e. they remained very fine in size and spherical in shape). Also, this type of carbide growth eliminated almost all preferential precipitation at martensite grain boundaries. The conventional martensite, on the other hand, had no existing 'sites' for carbide precipitation. The microstructure of these samples favoured the nucleation and growth of coarse carbide particles (at 600°C) with preferential precipitation at the martensite boundaries (intergranular fracture was observed in these specimens). This difference in carbide morphology and distribution could account for the difference in toughness between the two types of martensite at this stage of tempering.

At 600°C dislocation climb would take place and accordingly a certain degree of recovery was observed. However, recovery would be slower in the ausformed sample (relative to the conventional material) due to the greater restraining (i.e. pinning) effect of the finer, well dispersed carbides. The difference in dislocation density in conjunction with the difference in carbide morphology would explain why the ausformed martensite was not only much tougher, but also stronger than the conventional martensite.

The explanation for the improvement in the toughness of both types of martensite after tempering at 600°C (relative to that after tempering at 400°C) must take into account the absence of cementite plates in the matrix. At the lower tempering temperature (400°C) the large Fe_3C plates would be favourable sites for the initiation of low-energy, quasi-cleavage fracture. The effect of Fe_3C would be the same in both types of martensite, and this suggestion is supported by the absorbed impact energy results. The beneficial effect of the finer dispersion of particles in the ausformed sample (relative to the conventional martensite) became apparent only when the 'embrittling' effect of the Fe_3C plates was not present.

Although ausforming resulted in an overall refinement in the martensite structure, a very wide range of sizes was observed, and the dimensions of the largest martensite plates (and units) was substantially the same as that in the conventionally treated material. A parallel project at BISRA[17] has shown that the coarsest martensite plates have a dominant effect on the impact strength of the material. Therefore, it was felt that though the refinement in structure would bring about a slight improvement in the strength of the material,[18] it would not have any significant effect on the fracture toughness.

In summary, it can be said that an improvement in toughness was achieved by precipitating the chromium carbide in the right form (i.e. as fine, spherical particles) and in the right distribution (i.e. as a dense, *uniform* distribution with no preferential precipitation at grain boundaries).

CONCLUSIONS
1. The tempered martensitic structures examined in this work exhibited two types of fractures, quasi-cleavage and ductile rupture.

2. The presence of coarse plates of cementite promoted the initiation and propagation of quasi-cleavage.

3. Ausforming and subsequent tempering of the martensite promoted fracture by ductile rupture.

4. A finely dispersed alloy carbide phase was precipitated in the metastable austenite during the thermo-mechanical treatment. This uniform dispersion of carbide particles was 'inherited' by the martensite during the subsequent austenite→martensite transformation.

5. Ausforming improved the strength of the En56C steel without an accompanying drop in the toughness. This was true for the untempered specimens as well as for those tempered in the range 100°–400°C.

6. After tempering in the range 400°–600°C, the toughness of the ausformed martensite was substantially superior to that of the conventionally treated martensite. Also, the ausformed martensite had a higher tensile strength than the conventional material.

7. The improvement in mechanical properties (strength and toughness) of the ausformed martensite over that of the conventional martensite could be explained by the observed differences in carbide dispersion (and morphology) and dislocation distribution.

REFERENCES

1. S. V. RADCLIFFE and E. B. KULA: Proc. Conf. on Fundamentals of Deformation Processing, Sagamore, 1962.
2. S. J. MATAS et al.: AIME Proc. Fifth Mech. Working Conf., 1963, 143.
3. C. W. MARSCHALL: DMIC Report 192, 1963.
4. W. E. DUCKWORTH and P. R. TAYLOR: in 'Metallurgical developments in high-alloy steels', in ISI Spec. Rep. 86, 1964, 61–70.
5. G. THOMAS et al.: Proc. 2nd Int. Sym. on Materials, Berkeley, 1964.
6. W. E. DUCKWORTH et al.: JISI, 1964, 202, 135–142.
7. S. R. KEOWN and F. B. PICKERING: ibid., 1962, 200, 757–758.
8. W. BOLLMANN: Phys. Rev., 1956, 103, 1 588–1 589.
9. C. E. MORRIS: Met. Prog., 1949, 56, 696–699.
10. J. J. IRANI and S. GIBSON: BISRA Report, to be published.
11. J. SRAWLEY: NRL Research report.
12. C. D. BEACHEM and R. M. N. PELLOUX: ASTM Symp. on fracture toughness and its applications, Chicago, 1964.
13. B. YOUNG: M. Met. and Ph.D. theses, Univ. of Sheffield, 1959, 1961.
14. P. M. KELLY and J. NUTTING: Proc. Roy. Soc. 1960, 259A., 45–58.
15. P. M. KELLY and J. NUTTING: JISI, 1961, 197, 199–211.
16. J. J. IRANI: Ph.D. thesis, University of Sheffield, 1963.
17. R. PHILLIPS et al.: BISRA Report MG/E/111/64.
18. J. C. SHYNE et al.: Trans. ASM, 1960, 52, 346–375; and D. J. SCHMATZ et al.: Proc. NPL Symposium 'Relation between structure and mechanical properties of metals', 1963, 614.

Discussion 8

Chairman: Mr J. E. Russell (English Steel Corporation)

Mr D. Miles (University of Leeds) referred to work at Leeds on ausforming 12%Cr steels with secondary hardening additions. These were ausformed at 450°C on an Instrom testing machine.

Figure *A* was material given a 10% deformation. The test was stopped and an immediate relaxation occurred, as would be expected, but in time the stress increased and on reapplying the test there was a marked yield. This increased on successive stops after 5% deformations and there was much more marked yield and increase of contraction.

Figure *B* showed the same steel with most of the stress taken off before stopping the test. There was a much more marked contraction and on reapplying the deformation there was a much more marked yield. In fact, it was found that the amount of yield, or martensite burst, increased with the amount of contraction. This could be interpreted as carbon coming out of solution and settling on the dislocations while the test was being stopped. Carbon depletion would lower the M_s.

Calculations had been made and it was found that most of the carbon would have come out of solution during the stop. This might be a peculiar interpretation, but it was intended to repeat it, stopping the test for a much longer time.

If precipitation was likely to occur, it would seem that it would take place while dislocations were pinned between successive ausforming operations.

Mr A. I. Wildman (Bristol Aerojet Ltd) said that he wished to report the results of preliminary work done on ausforming an En40–C type steel. A tensile strength of 160 tons/in² had been obtained from a transverse specimen given 80% reduction at about 550°C, followed by air cooling and tempering at 300°C (an increase in tensile strength of 40 tons/in² over material conventionally heat treated). However, the steel was extremely brittle in this condition. There was no mention of transverse properties in Dr Irani's paper.

Dr J. J. Irani (BISRA, Sheffield) said that the transverse properties had not been measured at BISRA but there was a paper in an American Journal which showed that there was hardly any drop in the transverse direction. They had taken impact strengths from six different orientations in the bar and there was not much of a drop in ductility and the strength remained the same.

Mr Wildman pointed out that these results were on strip material and this was an instrumented bend test which had been used to measure the transverse ductility.

Dr P. M. Kelly (University of Leeds) asked if Dr Irani had used electron diffraction to identify the precipitates referred to in his paper as cementite and whether there was any electron diffraction evidence for the existence and identity of the small carbide precipitates in Dr Irani's ausformed material?

Professor J. Nutting (University of Leeds) said that Dr Irani had put forward the view that the good ductility of ausformed steels was associated with the fine precipitates. If this were the case then it should be asked why conventional vanadium-containing secondary hardening steels, in which after suitable heat treatment fine precipitates could be produced, did not show good ductility although they had a high strength. Did Dr Irani think that there were some other structural factors that differentiated ausformed and secondary hardened steels?

Dr Irani in reply to Dr Kelly agreed with him that the reproductions in the preprint did not show up the fine precipitates very clearly. He showed Dr Kelly the original micrographs and confirmed that the cementite plates were identified by electron diffraction. Because of the fineness of the particles of the second phase, clear electron diffraction patterns could not be obtained
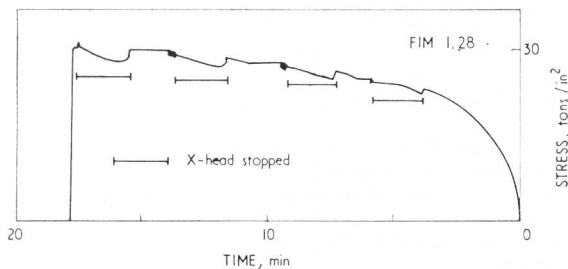
A *13%Cr–0·1%C–0·75%Mo steel; stress v. time while intermittently deforming metastable austenite in tension at 450°C; the stress increases when the test is stopped*
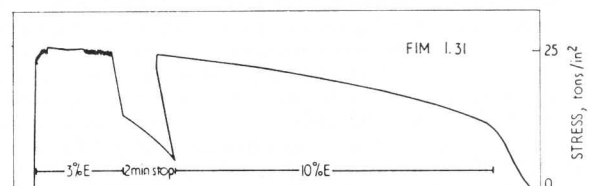
B *As Figure A, but the tension is relaxed before stopping the test; the increase of stress indicates a contraction*

from them. However, their presence was confirmed by dark field electron microscopy.

In reply to Professor Nutting, Dr Irani said that he had tried to deal with the point raised by Professor Nutting in his paper. He felt that since the ausformed martensite already had a uniform distribution of very fine precipitates, these would grow on subsequent tempering. This would result in a dense, uniform carbide precipitate with hardly any preferential precipitation at the martensite boundaries. On the other hand conventional martensites (with vanadium or any other carbide former) on tempering showed preferential precipitation at martensite 'unit' and 'bundle' boundaries. This difference in the distribution of the carbide precipitate could explain the differences in impact strength.

Dr R. H. Johnson (CEGB) said that he was very pleased to see how well the theory was substantiated by the work of Ankara and West. He pointed out, however, that he was now able to give a more detailed analysis than the one from which equation (1) was derived and this had been published recently.[1] The analysis had been extended to cover the case where the rate of transformation was slow and the weaker phase deformed by creep, as well as the case where transformation took place rapidly, as in quenching to martensite, where a yield stress was more applicable. An exact description of the mode of deformation was, for most cases, unimportant, since the equations derived in each of the extreme cases were essentially similar. The fuller analysis, however, indicated that the value of the numerical constant had changed very slightly from 2/3, as used by Ankara and West in equation (1), to a value of 5/6. This new value for A improved further the agreement in the examples quoted by Ankara and West where prediction and observation were compared, but this point should not be stressed too highly because of the approximations still required in the theory and the necessity of using approximate values for $\Delta V/V$ and the yield stress of the austenite.

He was also interested to note that Ankara and West's results indicated a linear relationship between applied stress and extension which was valid to the limit where the applied stress approached the yield stress. This feature substantiated his view[2] that any suggestion of a non-linear relationship between applied stress and transformation strain would require reappraisal at low applied stresses. It was possible to show theoretically that linearity up to stresses of the same order as the tensile yield stress would be expected in the case of deformation by yielding. For slower rates of transformation both theory and experiment indicated that linearity only held for values of the applied stress up to about half the flow stress.

Dr D. R. F. West (Imperial College of Science and Technology) agreed with regard to the comparison between experiment and theory, that the experimental accuracy of the various parameters used did not enable the constant in the equation to be evaluated precisely from the results; he and Dr Ankara would have been satisfied with a less close agreement between the predicted and experimental values. Regarding the extent of the linearity, here again the experimental results did not locate exactly the point at which the data departed from linearity, but a significant departure from linearity corresponded closely to the yield stress.

Mr J. J. de Jong said that in one of his experiments, Dr West referred to an austenitizing time of 1 second at 850°C. How did he manage? Did he bring his specimen into temperature equilibrium just beneath the transformation temperature, which was probably not a sharp transformation temperature but rather a transformation region? Under these circumstances, one would not expect the ferrite and carbides to have completely transformed into austenite in such short times as 1 second. He asked if Dr West had made any metallographic investigations on his specimens just after quenching. He drew attention in this respect to some micrographs which were published a few years ago by a German investigator, Dr Rose,[3] who made micrographs of

specimens after a very short austenitizing time, and in which one could easily see that carbides were on their way to being dissolved but were not quite dissolved yet. One could clearly distinguish areas which had been transformed and areas which had not yet been transformed. He had been able to confirm these results on optical and electron micrographs. He thought that the structure obtained after 1 second of austenitizing at 850°C would be quite different from that in other austenitizing conditions in the experiments.

He was therefore rather surprised that the points under those conditions fitted so well in the curves. He asked for Dr West's comments.

Dr West said that the specimens were rapidly heated to the austenitizing temperature. In short-time treatments the average heating rate was about 200 degC/s and the time spent in the austenite range was estimated accordingly; 1 second was a reasonable estimate for the time at 850°C in the shortest treatment. In agreement with Mr de Jong's comment, the austenite was not expected to be homogeneous after this time. However, since the samples were rapidly heated from the martensitic condition, little partitioning of alloy elements would be expected during any tempering that occurred, and the process of austenitizing might be expected to involve essentially the solution of fine carbides and the diffusion of carbon.

The extent of carbide solution and austenite homogeneity were not determined, but an experiment was done by Dr A. S. Sastri[4] in which a plain carbon eutectoid steel, initially in the martensitic condition, was heated to 850°C in a total time of 4 seconds followed by immediate air cooling. The resultant structure as observed by transmission electron microscopy was essentially pearlitic; although a few undissolved carbide particles may have been present, it appeared that a good degree of homogenization had been achieved in the austenite.

Professor H. W. Paxton (Carnegie Institute of Technology) said that the yield stress when martensite was forming was surely that which was relevant to quite high strain rate. He did not know how strain rate sensitive austenite was. He wondered if the authors had any feelings on this.

Dr West said that this was a question which Dr Johnson could probably answer better than himself. Dr Johnson had dealt with the effect of strain rate in the recent paper to the Royal Society; the strain rate in the present work was relatively high.

Dr Johnson said that strain rate sensitivity could be important but was allowed for in the analysis, for example, by incorporating the stress exponent for steady-state creep in the case where the strain rate varied as some power of the stress. This effect, however, was difficult to study experimentally since the overall rate of straining estimated from the time for complete transformation was not necessarily the same as the strain rate in the local regions adjacent to the interface between transformed and untransformed material.

Mr M. A. P. Dewey (Aeon Laboratories) returned to the discussion between Professor Nutting and Dr Irani and suggested that in these materials the ductile/brittle transition temperature was just as important as the room temperature ductility. He wondered if anyone had compared the transition temperature of a steel in the ausformed and hot rolled conditions.

Dr Irani replied that recent tests on an 3% Cr–0·4% C steel indicated that ausforming did not affect the transition temperature. He pointed out that ausformed (and subsequently tempered) material did not show a 'sharp' transition temperature, but rather a gradual fall in impact energy with decreasing test temperatures.

Mr R. Phillips (Colvilles Ltd) said that Dr Irani stated: 'The need to nucleate a larger number of microvoids to propagate

the fracture would also explain qualitatively the higher absorbed impact energy of the ausformed martensite relative to the conventional martensite'.

He would suggest that the observation of two features simultaneously did not necessarily mean that they had a cause and effect relationship and that it was not obvious that the larger number of microvoids was the cause of the higher absorbed impact energy.

Instrumented Charpy tests appeared to show that the bulk of the energy absorbed in ductile failure was due to plastic strain rather than the actual fracture. It might also be noted that, to a first approximation, the amount of surface created in a ductile fracture and therefore the surface energy created was independent of the dimple size. The dimples could only be important in determining the energy absorbed in fracture if the bulk of the plastic strain was used to create the dimples. It was not clear that this was the case, but one might, nevertheless, investigate the effect which change of dimple size would have if the bulk of the plastic strain were used to create them.

Unless the finer dimples were at least as deep as the coarse ones, the applied stress would move through a smaller distance in creating the fracture with the finer dimples. With any reasonable dependence of the stress on strain, less energy might be expected to be dissipated in propagating the fracture with the finer dimple spacings unless the required stress was very much larger for this fracture. This approach might be elaborated a little.

The absorbed energy might be written as

$$N \times S \times E \dots\dots\dots\dots\dots\dots\dots\dots\dots\dots\dots\dots (1)$$

where N was the number of dimples per unit area, S the average stress acting during their formation, and E the strain required to produce them. If the voids were spherical and the strain per void was proportional to its volume, as seemed to follow from the model of ductile fracture proposed by McLean[5] E would be proportional to $N^{-3/2}$ and equation (1) would become

$$N^{-\frac{1}{2}}S$$

Thus the larger N and the finer the dimples the smaller the energy absorbed. As the power dependence of the strain per void on N varied from $-3/2$ towards zero, $N \times S \times E$ became less and less dependent on N, until when it was -1 (i.e. the strain was proportional to the surface area of the microvoid) there was no net dependence of the absorbed energy on N. This was the smallest rate of dependence of the strain per void on N that could reasonably be postulated and under these conditions, this simple model indicated that, if all the energy absorbed was due to the creation of microvoids, then it was a simple linear function of the average stress acting during void formation. It could perhaps be suggested that this was the explanation of Dr Irani's result. However, he was not trying to suggest that void formation was the major factor in determining the energy absorbed in ductile fracture or that the above models should be seriously applied to this problem, but merely that Dr Irani's contention was very far from obvious and indeed in a simple analysis appeared wrong. If he had a more sophisticated model which supported his view it would be interesting to have it described.

Mr Fussell and he had worked on controlled rapid reausten-tizing of En30B at BISRA and were able to compare the ductile fracture in Charpy tests of specimens with 34 and 7 μm prior austenite grain sizes. The fracture surface of the material with the 7 μm grain size contained five times as many dimples per unit area as that of the 34 μm grain. In this case, however, the finer dimple structure was associated with a lower absorbed energy despite the 7 μm grain size steel being stronger (14 tons/in² on the proof stress and 8 tons/in² on the UTS) and the reductions in area being the same, although the elongation was lower (16% compared with 19%). The full transition curves from this work were shown in Figure C and the results were of course at variance with Dr Irani's suggested dependence of energy absorbed in ductile fracture on microvoid spacing. He would suggest, therefore, that the higher absorbed energy in
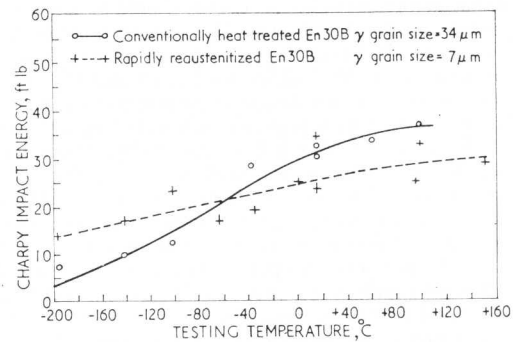
C *Charpy impact energy transition curves; En30B tempered at* 200 °C *for 1 h*

ductile fracture in the ausformed sample was due to the effect of the different distribution of precipitates and dislocations on the energy absorbed in plastic strain before fracture rather than to the finer dimples *per se*. The point might also be made that the effect of ausforming on the Charpy transition curve was very sensitive to the exact processing conditions. Dr Irani had shown the importance of tempering temperature in this respect, and Figure D from Kula and Dhosi[6] showed the importance of amount and temperature of aus-deformation. As the strengthening response also varied in a complex manner with these parameters, detailed information on both strength and toughness was required for ranges of values of these parameters if the optimum ausforming and tempering treatments for specific applications were to be evaluated.

Dr Irani, in reply to Mr Phillips said that he had not tried to convey the impression that the size of the fracture units *per se* was a measure of the impact energy. The point he had tried to make in his paper was that in the *same* ausformed steel under the *same* testing conditions, a substantial increase in impact absorbed energy was accompanied by a refinement of the 'dimple' size seen on the fracture surface. This observation was supported by experimental results from several ausformed and conventionally treated steels.

He thought the ductile cracks could be nucleated at the precipitate/matrix interface. Though fractographic work showed that there was no evidence that each precipitate particle would nucleate a crack, it was fair to say that, statistically, the greater the number of precipitate particles the greater the number of crack nuclei.

He did not agree with Mr Phillips' suggested analysis which
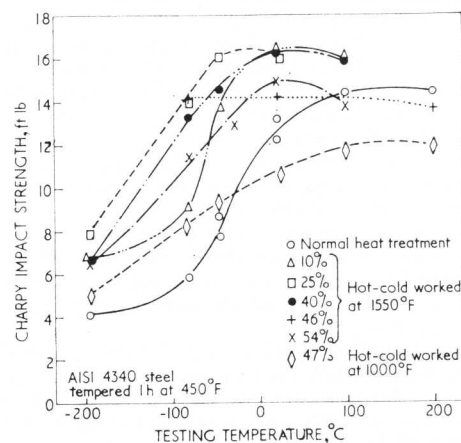
D *Effect of hot–cold working on impact toughness and transition temperature of AISI* 4340 *steel.*

showed that an increase in impact energy should be accompanied by an increase in the dimple size. This was contrary to experimental facts. He had had the opportunity of examining the work mentioned by Mr Phillips (of Fussell and Phillips) and would like to point out the danger of putting too much reliance on differences in fracture topography when the impact absorbed energy differed by only 6–8 ft lb or less. In his experience this difference in impact energy was well within the limits of experimental scatter observed in Charpy impact tests. His own observations on fracture topography were based on specimens which showed a difference of over 60 ft lb in impact absorbed energy (e.g. Figure 1D of his paper). Furthermore, it was difficult to control rigidly the heating cycles used in Mr Phillips' work (on rapid reaustenitizing), and differences in the state of the parent lattice could account for the observed differences in fracture topography.

Similar work (as reported in his own paper) had been done on the material (En30B) examined by Fussell and Phillips. Two results from the ausformed material were shown in Figures E and F.

Both Figures E and F were replicas of fracture surfaces of specimens tempered in the low-temperature range after ausforming. The specimen shown in Figure E (coarse dimples) gave an impact energy of 30 ft lb compared with nearly 55 ft lb absorbed by the specimen shown in Figure F (finer dimples).

He agreed with Mr Phillips that the dislocation configuration and precipitate distribution would affect the nucleation and growth of cracks. In fact, the main theme of his paper had been to show the variation in toughness with precipitate distribution and dislocation configuration.

It was very true that processing conditions (during ausforming) governed the ductile/brittle transition, and detailed work was in hand to evaluate the effect of variations in important operational parameters on the strength and toughness of the ausformed product.

Dr T. Boniszewski (BWRA) said that in the controversy between Dr Irani and Mr Phillips, he would support Mr Phillips. In the ductile fracture one could often observe fine-dimpled surface when the energy absorption was low. High-energy absorption was observed in the ductile fracture when large amounts of plastic deformation preceded final separation. However, when the microstructure was already densely packed with dislocations, low-energy ductile fracture might show a fine dimpled surface. 18% Ni–Co–Mo maraging steel provided an example of low-energy (20–40 ft lb) ductile fracture with fine surface dimples. However, various low-carbon low-alloy structural steels could have high-energy absorption (in excess of 100 ft lb), but their surface dimples were coarser than those of the maraging steel. The maraging steel had a very high dislocation density, and such a state was equivalent to cold worked material.

The different capacity of various steels for energy absorption could be inferred from the shape of their stress–strain curves obtained in slow tension (Figure G) on a 'hard' machine. A steel which would work harden before necking would show high-energy absorption in ductile fracture under impact (curve A in Figure G). A steel which was already work hardened, or had high density of transformation dislocations (curve B, Figure G), would absorb little energy in ductile fracture under impact. The two steels might show similar total elongations and similar reductions of area. However, it was the uniform elongation, before the onset of plastic instability, that mattered in energy absorption under impact. In maraging steel, there was no more space for new dislocations and the steel began to neck immediately after it had yielded.

These considerations should be regarded only as contributory factors and not the whole explanation. There were other factors which affected the ductile fracture, e.g. the form and distribution of inclusions in steel. The same volume fraction of inclusions, finely dispersed along grain boundaries, could give fine dimples and low energy absorption, while large inclusions would give large dimples and high energy absorption.
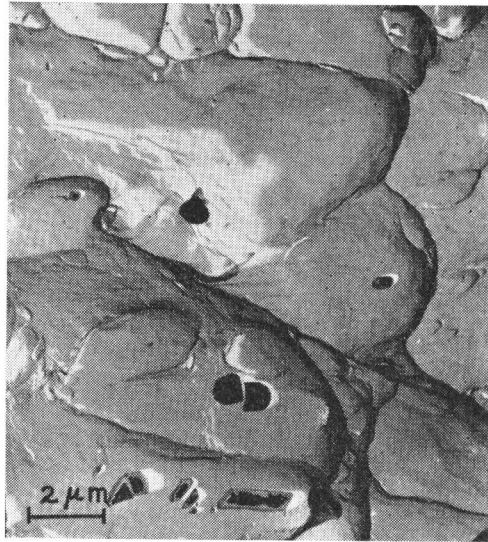
E   *Ausformed En30B steel; coarse dimples*   ×5000

In certain cases the effects of work hardening and inclusion distribution might work in opposite directions. Cold work might upset grain boundary networks and thus increase the energy absorption from one source while lowering it by increasing the dislocation density.

The obvious conclusion was that more systematic work needed to be done at universities to study all the aspects of the ductile fracture in metals.

Dr Irani replied that comparisons with maraging steels were not relevant. Of course, one could expect to see examples when a fine ductile fracture was observed in a specimen which showed a low impact energy. He would like to emphasize the point he had made in reply to the last question that the size of the fracture unit *by itself* was not a measure of impact energy. However, for specimens of the *same* steel tested under identical conditions, a *substantial* increase in absorbed impact energy was accompanied by a decrease in the size of the fracture units (i.e. 'dimples'). One must be careful when comparing the fracture topography of maraging steels (with very low carbon and a high proportion of alloying elements) with that of ausforming steels (with 0·3–0·4% C and a much lower alloy content).

Mr F. B. Pickering (The United Steel Cos. Ltd) said that he tended to agree with Dr Boniszewski and Mr Phillips who made the point during the discussion of the paper by Dr Irani that a finer ductile dimple size did not necessarily mean a higher energy absorption during impact testing, and would simply quote the case of pearlite which could, in ductile fracture, give a very fine dimple size and yet had a very much lower ductile impact energy than had ferrite, which had a coarse dimple size.

The ausformed specimens were apparently ductile, while the conventionally treated specimens were brittle. Had Dr Irani examined the fractures of ausformed and conventionally treated material so as to obtain full impact transition temperature curves and show the change in transition temperature caused by ausforming? A true comparison of ductile and brittle fractures could then be made.

Dr Irani said that in replies to Mr Phillips and Dr Boniszewski he had made it clear that the dimple size had not been considered as an absolute measure of impact absorbed energy. Of course, a difference in structure would lead to a marked change in the fracture topography. The initiation and propagation of micro-voids in ferrite was entirely different from that in pearlite and this would certainly be reflected in the relative sizes of the
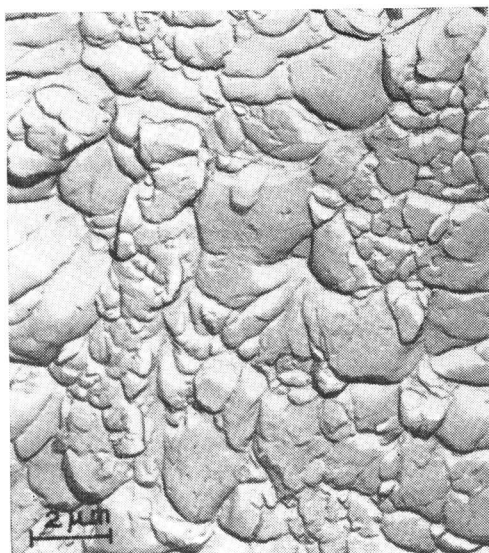
F *Ausformed En30B steel*; *fine dimples* ×5000

A conventional structural steel
B 18% Ni–Co–Mo maraging steel
G *Schematic representation of nominal stress–strain curves*

dimples observed on the fracture replica. The work described in the paper dealt with the same basic structure. He would like to refer Mr Pickering to Figures *E* and *F*.

Work was in hand to establish full impact 'transition curves' (see the answer to Mr Dewey) for ausformed and conventionally treated martensites.

Professor Paxton, adding to what Mr Pickering had said, thought that it might now be relevant to start looking at values of the various fracture toughness parameters for these materials.

Mr J. H. R. Page (English Electric Co. Ltd) said that Dr Irani had mentioned that he had found no change in transition temperature by comparing conventionally heat treated materials and ausformed materials. Recently[7] in a copy of *Nature*, Dr McLean postulated that since the final grain size of ausformed materials was much smaller than that of conventionally heat treated materials this should lead to a lowering of some 20–30 degC in the ductile/brittle transition temperature. He wondered if Dr Irani would comment.

Dr Irani said that he was aware of the work of Dr McLean, which was done in close cooperation with BISRA. However, further work on various steels indicated that ausforming did not have an appreciable effect on grain-size refinement. This was in agreement with the work of Zackay and his co-workers at Fords in America. He also pointed out the information he had given in his reply to the last question – that ausformed material did not show a sharp 'transition' temperature, in the accepted sense of the term.

Dr K. J. Irvine (The United Steel Cos. Ltd), summarizing the proceedings of the Conference, said that it was helpful to see to what extent the objects of the Conference had been achieved. One of the objects was to bring together those university staffs working on some of the basic questions of ferrous metallurgy and industrial staffs working on the same problem and to get close and detailed contact between them. From this point of view, it could be claimed that the Conference had been successful. If one looked at the membership list, there was an extremely good representation from the universities. He had counted over 60 university staff coming from 9 university departments and many technical colleges. On the industrial side most of the leading steel companies were represented. He wanted to make a
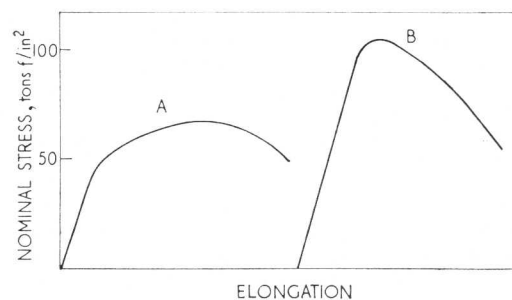
particular point about foreign representation. As far as these BISRA Conferences went, this had been the best in overseas representation. There were representatives from many countries, who had made important contributions both to the papers and the discussions, and these had been greatly valued.

In his opening paper, Dr Christian had said that the crystallography of martensite was based on simple ideas but was usually presented in an abstruse mathematical form and, therefore, seemed rather removed from the practical work of steelworks. This was very true and it could be asked whether this Conference had changed the position. Personally, he had made rather heavy weather of the thermodynamics and kinetics of martensite. Looking for a simple rationalized theory, it was clear that there was some way to go before this was achieved.

Turning to the structure of martensite, he thought that it could be claimed that real progress had been made. Very elegant techniques had been developed, particularly in thin-film electron microscopy, and these were now giving a very good picture of the martensitic structure. With regard to strength he was very encouraged. One important objective was to try and get a quantitative assessment of strength and he thought it was encouraging to see the progress being made with the quantitative assessment of the factors contributing to the strength of martensite. There might be some difference of opinion about the relative order of these effects but he thought there was agreement developing about what these effects were. Once the factors contributing to strength were known, then it would be possible to exploit them to improve properties. It was clear that one of the important factors was the fineness of the structure and it was from this that one was led to mechanical working processes or rapid heat treatment processes to obtain improved strength with adequate ductility. It might seem to the academic representatives present that this point about ductility was being needlessly represented, but as far as users were concerned increased strength was useless without accompanying ductility and toughness. It seemed likely that the next year or two would see considerable progress in these special treatments applied to the martensitic structure.

One aspect which he had found particularly interesting related to the suggestion coming from Professor Nutting and Dr Kelly about aging effects in martensite. He gathered that Professor Nutting was saying that aging made a major contribution to the strength of martensite and that, if this aging reaction could be more closely controlled, it might be possible to obtain improved properties. This made good sense to him because his experience of metallurgical structures told him that he must be careful when dealing with aging reactions. Uncontrolled aging reactions were a frequent cause of poor properties in steel. He would, therefore, encourage Professor Nutting in this work and hope to hear of progress at a later date.

Turning to bainite, it was also true to say that there was very encouraging progress. The same techniques of thin-film electron microscopy had advanced the understanding of this structure

very much. From diffraction work it was possible to get a definite picture of this transformation and there was now reasonable understanding of the bainitic transformation over a range of carbon content from 0·1 to 1%. There was still some uncertainty about the upper and lower bainite reaction in low-carbon steels of the order of 0·1%C and, since such steels were of considerable commercial importance, he would encourage further work on this point.

Summarizing, therefore, with regard to martensite and bainite, which were two important structures in ferrous metallurgy, he thought that the Conference could show that considerable progress had been made. He would say, however, that they were not the most important structures in ferrous metallurgy and as he looked at the large group of academic research staff present he could realize the tremendous potential of such a group working on other topics in ferrous metallurgy. In his opinion, the most important structures were ferrite and pearlite. At the Conference BISRA held two years ago on plain carbon steels the university representation was very small and there were apparently few people working in this field. It was difficult to understand the preference for work on martensite compared to ferrite and pearlite, and if he could encourage some of the people present to switch their research effort to this field then this could be regarded as another benefit arising from the Conference.

In closing the Conference the Chairman thanked Professor Owen and the Organizing Committee, those who had presented papers, those who had contributed to the discussion, the Vice Chairmen, Professor Honeycombe and Dr Broom, and those responsible for the Conference arrangements.

A vote of thanks to the Chairman was proposed by Mr Duckworth and carried by acclamation. The Conference then ended.

REFERENCES

1. G. W. Greenwood and R. H. Johnson: *Proc. Roy. Soc.*, 1965, **283A**, 403–422.
2. R. H. Johnson: *Acta Met.*, 1965, **13**, 441–443.
3. A. Rose: *Härt. Techn. Mitts.*, 1958, **13**, 46–76.
4. A. S. Sastri and D. R. F. West: *JISI*, 1965, **203**, 138–145.
5. D. McLean: 'Mechanical properties of metals' 235; 1962, New York–London, Wiley.
6. E. B. Kula and J. M. Dhosi: *Trans. ASM*, 1960, **52**, 321.
7. D. McLean: *Nature*, 1965, **205**, 6 March, 1000–1001.

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