Evaluation of Toughness in AISI 4340 Alloy Steel Austenitized at Low and High Temperatures

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It has been reported for as-quenched AISI 4340 steel that high temperature austenitizing treatments at 1200°C, instead of conventional heat-treatment at 870°C, result in a two-fold *increase* in fracture toughness, K_{Ic}, but a *decrease* in Charpy impact energy. This paper seeks to find an explanation for this discrepancy in Charpy and fracture toughness data in terms of the difference between K_{Ic} and impact tests. It is shown that the observed behavior is independent of shear lip energy and strain rate effects, but can be rationalized in terms of the differing response of the structure produced by each austenitizing treatment to the influence of notch root radius on toughness. The microstructural factors which affect this behavior are discussed. Based on these and other observations, it is considered that the use of high temperature austenitizing be questioned as a practical heat-treatment procedure for ultrahigh strength, low alloy steels. Finally, it is suggested that evaluation of material toughness should not be based solely on K_{Ic} or Charpy impact energy values alone; both sharp crack fracture toughness and rounded notch impact energy tests are required.

RECENT research[1] into the influence of austenitizing treatment on the mechanical properties of as-quenched (untempered) AISI 4340 steel has shown that compared to conventional austenitizing at 870°C, increasing the austenitizing temperature to 1200°C can raise the plane strain fracture toughness (K_{Ic}) by a factor of two (without reduction in yield strength). However, concurrent with this *improvement* in K_{Ic} there is an unexplained and perplexing *reduction* in the Charpy impact energy. The same effect has been reported for British Ni-Cr-Mo and Ni-Cr-Mo-V steels,[2] 4340[3,4] and 300M[3] steels, in both as-quenched and quenched and tempered (<350°C) conditions. Since K_{Ic} and the Charpy impact energy are both measures of material toughness, it appears paradoxical that *high* temperature austenitizing gives the greater toughness when rated by K_{Ic}, while *low* temperature austenitizing gives the greater toughness when rated by Charpy impact energy. In the face of this conflicting evaluation of toughness, it is tempting for the scientist to simply adopt the results of the K_{Ic} test and disregard the results of the Charpy test as "inadequate", since it is no doubt true that the K_{Ic} test is based on a theoretical foundation of greater solidity than is the Charpy impact test. In reality, however, the contradictory results of the two techniques for measuring toughness appear to be a real phenomenon and are indeed an indication of important differences between fracture induced by sharp (K_{Ic}) and blunt (Charpy) notches. These differences may be summarized as follows: 1) the Charpy test measures the energy required to cause complete failure of the specimen and therefore will include a contribution from plane stress, shear lip formation. The fracture toughness test, on the other hand, measures a critical value of the stress intensity (K) at the crack tip necessary to cause plane strain unstable fracture. In a test-piece of valid thickness,[5] this value will be virtually independent of shear lip formation. 2) The strain rate in an impact test is several orders of magnitude greater than in a K_{Ic} test. In fact, when expressed in terms of rate of increase in stress intensity at the notch tip ($\dot{K}$), $\dot{K}$ for impact ("dynamic") testing is of the order of 10^5-10^6 MPa$\sqrt{m}$/s compared with <3 MPa$\sqrt{m}$/s for ("static") K_{Ic} tests. 3) There is a marked difference in the root radius (ρ) of the stress concentrator in the two tests. Charpy test-pieces contain a V-notch ($\rho \simeq 0.25$ mm), whereas K_{Ic} test-pieces contain a fatigue precrack ($\rho \rightarrow 0$).

It is the purpose of this paper to offer an explanation of the contradictory toughness results reported for the 4340 alloy, in the light of these basic differences, with the belief that this explanation may have considerably broader application.

EXPERIMENTAL PROCEDURE

The material used in the investigation was aircraft quality (vacuum-arc remelted) AISI 4340 hot-rolled bar, received in the fully annealed condition, and having the following composition (wt pct):

C	Mn	Cr	Ni	Mo	Si	S	P	Cu
0.40	0.80	0.72	1.65	0.24	0.24	0.010	0.004	0.19

Two heat-treatments were studied: 1) the conventional austenitizing treatment of 1 h at 870°C, followed by an oil quench and 2) the heat-treatment, recommended by Lai *et al*,[1] of 1 h at 1200°C, followed by a salt quench to 870°C for 1/2 h, then oil quenching to room temperature. The structures obtained by these heat-treatments are hereafter referred to as the 870°C and the 1200-870°C structures respectively.

The ambient temperature (22°C) uniaxial tensile

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Table I. Ambient Temperature Longitudinal Tensile Properties for Alloy AISI 4340 in the Oil Quenched Condition (after Lai *et al*[1])

Austenitizing Temperature, °C	0.2 Pct Proof Stress		Ultimate Tensile Strength		Elongation, Pct	Reduction in Area, Pct	True Fracture Strain	Prior Austenite Grain Size, μm
	ksi	MPa	ksi	MPa				
1200-870	231	1593	318	2193	3.2	7.8	0.07	254 to 360
870	231	1593	322	2217	9.0	30.8	0.18	24 to 32

properties and prior austenite grain sizes for these structures are shown in Table I. Toughness properties were measured using Charpy V-notch impact tests, plane strain ("static") fracture toughness (K_{Ic}) tests, and dynamic fracture toughness (K_{Id}) tests. Charpy impact energies were determined using standard sized (10 mm sq) ASTM Charpy V-notch specimens,[6] broken in a pendulum type impact machine of hammer velocity 3.3 m/s. K_{Ic} values were determined in accordance with ASTM specifications,[5] using 15.4 mm thick CTS test-pieces. All K_{Ic} values were found to be "valid" according to this specification. K_{Id} data were determined using fatigue precracked standard Charpy specimens[7] broken using an instrumented (Dynatup) Charpy impact machine at a hammer velocity of 1.36 m/s. The complete experimental procedure for dynamic K_{Id} testing has been described elsewhere.[8] Test-pieces for all experiments were machined in the longitudinal ($L-T$) orientation.

Table II. Ambient Temperature Longitudinal Toughness Properties for Alloy AISI 4340 in the Oil Quenched Condition

Austenitizing Temperature, °C	Charpy V-Notch Impact Energy		Plane Strain Fracture Toughness, K_{Ic}		Dynamic Fracture Toughness, K_{Id}	
	ft lb	J	ksi$\sqrt{in}$	MPa$\sqrt{m}$	ksi$\sqrt{in}$	MPa$\sqrt{m}$
1200-870	5.4*	7.32	63.8*	70.1	54.2	59.6
	6.0*	8.13	66.6*	73.2	48.8	53.7
	5.0	6.78	61.2*	67.2		
	2.0	2.71				
	6.0	8.13				
870	7.5*	10.20	31.1*	34.2	36.6	40.3
	6.1*	8.27	39.0*	42.9	30.0	33.0
	7.5	10.20	32.3*	35.5	33.3	36.6

*After Lai *et al.*[1]

RESULTS

The ambient temperature (22°C) toughness properties for both structures are shown in Table II. Two important features are shown by these data: 1) As originally reported by Lai *et al*,[1] the 1200-870°C structure has *superior* K_{Ic} but *inferior* Charpy values, with respect to the 870°C structure. The difference in Charpy values is, however, small. 2) Both K_{Ic} and K_{Id} results show the 1200-870°C structure to be the tougher, although there is some reduction in the dynamic values due to the higher strain rate.

In the following paragraphs we now consider these results in terms of the basic differences between the Charpy and fracture toughness tests, as summarized above.

1) Shear Lip Effects

It is well known that the shear lip portion of the fractured surface of a Charpy test-piece requires more energy per unit area to fracture than does the flat portion. Therefore, a significant increase in the proportion of plane stress shear lips in the 870°C structure, over that of the 1200-870°C structure, could possibly account for the small improvement in Charpy energy of the 870°C structure.* Examination of the broken Charpy test-pieces, however, confirmed that there was no measurable difference in the shear lip area for the two structures. In fact, the proportion of shear lip area for both structures was always less than 4 pct. It is concluded that fractures in both Charpy and K_{Ic} tests occurred under fully plane strain conditions, and therefore the contribution from shear lip formation can be eliminated as an explanation for the difference in Charpy and K_{Ic} test results.

*In view of the fact that the yield strengths of the two structures were identical, and that the K_{Ic} for the 870°C structure was less than for the 1200-870°C structure, one would *not* expect such an increase.

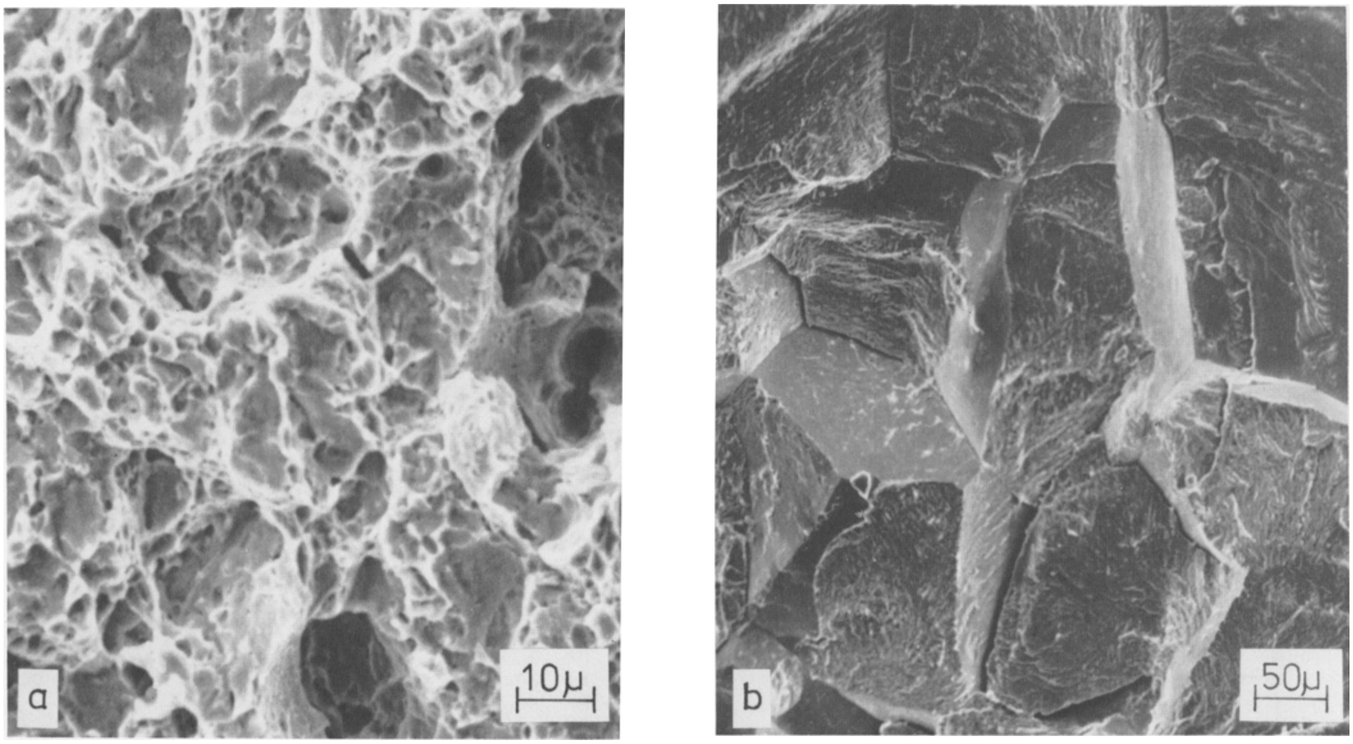
2) Strain Rate Effects

There is always a possibility when comparing Charpy impact and K_{Ic} values that the higher strain rate of the impact test ($\sim 10^1 - 10^2$ s^{-1} compared to $\sim 10^{-4}$ s^{-1} in static tests) may induce a change in fracture mechanism through an increase in the ductile/brittle transition temperature.

In the present results, it might be expected that the transition temperature would be higher for the 1200-870°C structure because of the larger grain size. Thus, increased strain rate may shift the transition temperature of the 1200-870°C structure above room temperature in the Charpy test, whilst leaving the transition temperature of the 870°C structure below room temperature. This would result in a low Charpy impact energy for the 1200-870°C structure.

Scanning electron microscopy of fracture surfaces of both Charpy and K_{Ic} test-pieces, however, showed that the fracture mechanisms were identical in both tests for a given structure. Failure in both Charpy and K_{Ic} test-pieces occurred by quasicleavage and fibrous rupture in the 870°C structure, and by intergranular and fibrous fracture in the 1200-870°C structure (Fig. 1). Since for a given heat-treatment the fracture mechanisms, at both high and low strain rates, are the same, change of mechanism can be eliminated as an explanation for the contradictory toughness results.

There is an additional possibility that a strain rate effect could occur without change in fracture mecha-



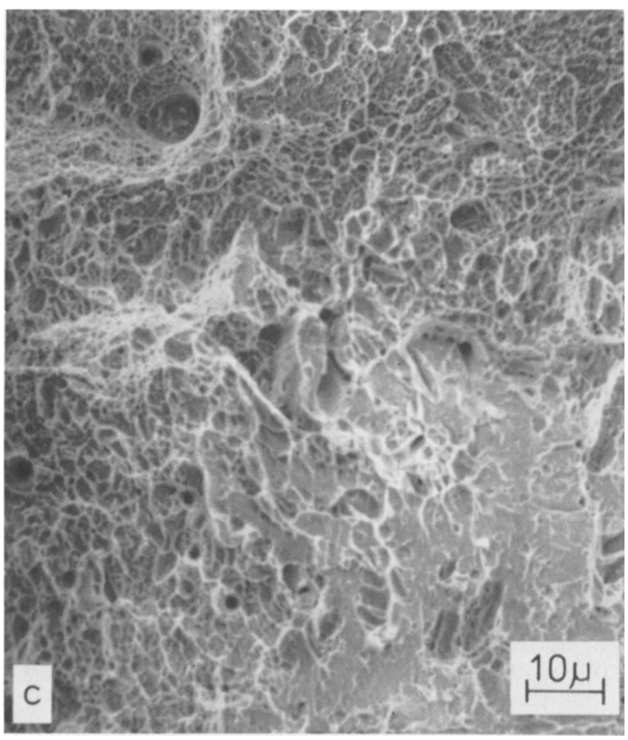


Fig. 1—Fractographs of as-quenched 4340; showing (*a*) in structure austenitized at 870°C, quasicleavage (~ 65 pct) linked by areas of fibrous rupture; and (*b*) and (*c*) in structure austenitized at 1200-870°C, intergranular cracking (~ 55 pct) and transgranular fibrous rupture. Several intergranular facets show evidence of microvoid coalescence on grain boundary surfaces (*c*). *Note:* 1200-870°C structure fails partly by intergranular fracture, and not, as erroneously reported in Ref. 1, by 100 pct fibrous rupture.

nism, through, for example, an effect on deformation flow properties. To test this possibility dynamic fracture toughness (K_{Id}) tests were conducted on precracked test-pieces. The results are listed in Table II, where it can be seen that, although there was some reduction in toughness for the higher austenitizing treatment, the 1200-870°C structure still shows superior fracture toughness compared to the 870°C structure. Since this result occurs at the same strain rate as in standard Charpy tests, which show the opposite result in terms of which structure is tougher, it is clear that strain rate sensitivity does not provide an explanation for the contradictory toughness results.

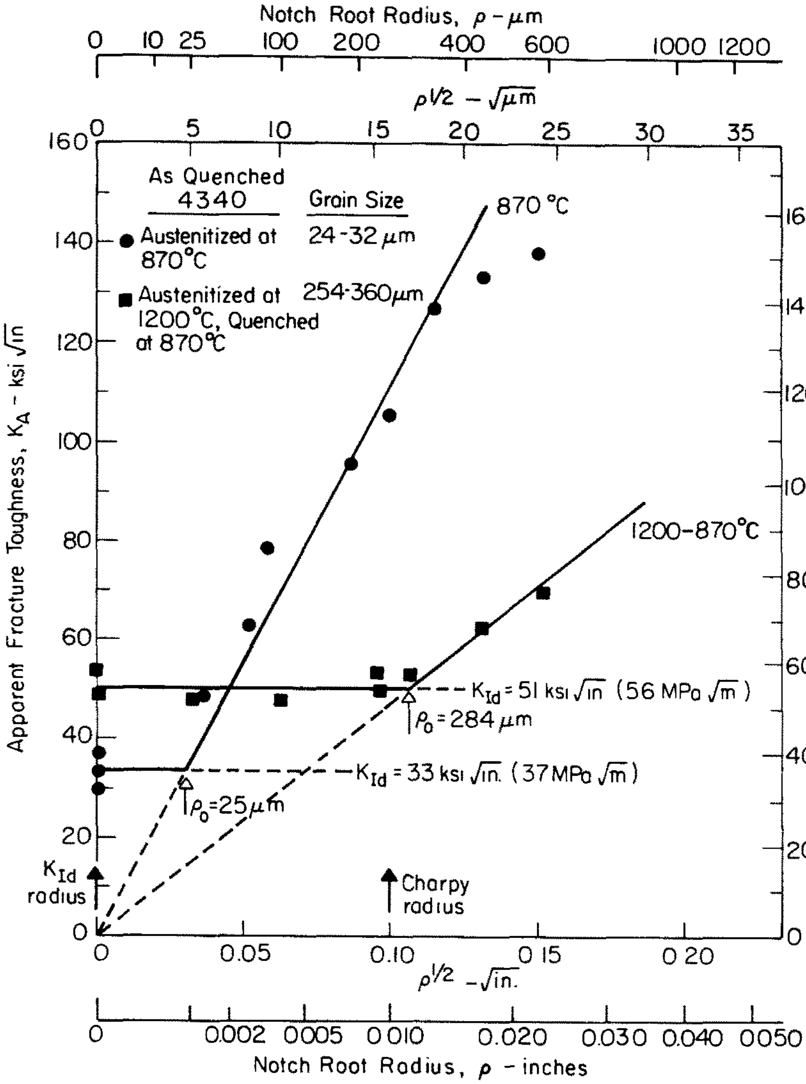


Fig. 2—The relationship between toughness, measured by the apparent dynamic fracture toughness (K_A) from instrumented Charpy tests, and notch root radius (ρ) in oil quenched AISI 4340, for the 1200-870°C and 870°C structures. ρ_0 is the "limiting" root radius, K_{Id} the dynamic fracture toughness.

3) Notch Root Radius Effects

A significant difference between Charpy and fracture toughness testing is the radius of the stress concentrator introduced into the test-pieces. Charpy test-pieces contain a V-notch (ρ = 0.25 mm) whereas K_{Ic} test-pieces contain a fatigue precrack ($\rho \rightarrow 0$). To determine how measured toughness varies with notch root radius for each austenitizing treatment, a series of Charpy test-pieces was prepared with notch root radii ranging from a fatigue precrack to a 0.58 mm radius V-notch. The specimens were broken at room temperature (22°C) in an instrumented Charpy machine, and impact energy and "apparent" dynamic fracture toughness, K_A,* measured in each case (Table III). In Fig. 2 the apparent fracture toughness, K_A, is shown as a function of the square root of the notch radius, $\rho^{1/2}$, for both microstructures. The important feature of this figure is that for small root radii (ρ < 0.05 mm) the toughness of the 1200-870°C structure exceeds that of the 870°C structure, whereas at larger radii (ρ > 0.05 mm) the reverse is the case. A similar effect has been observed for fine and coarse grained 4340 tested at liquid nitrogen temperatures.[12] Thus, it can be seen that in fracture toughness tests, where $\rho \rightarrow 0$, the 1200-870°C structure will have the higher K_{Ic} value: but for Charpy tests, where ρ = 0.25 mm, the 870°C structure will fracture at the larger K_A value, and thus show the larger impact energy. This is precisely the behavior reported previously.[1-4]

*"Apparent" fracture toughness refers to the value of K_{Ic} (or K_{Id}) measured ahead of a rounded notch of radius ρ, and is generally used as a means of estimating the fracture toughness without recourse to fatigue precracking.[9-11]

An explanation for this effect can be deduced by examining the simple theory, developed by Tetelman and coworkers,[9,10] to account for the influence of notch root radius (ρ) on fracture toughness. For constant stress-controlled fracture under elastic-perfectly plastic conditions, it was proposed that failure occurs when the maximum tensile stress, σ_{yy}^{max}, located at the plastic-elastic interface exceeds a critical fracture stress, σ_F, for failure. Thus if

$$\sigma_{yy}^{max} = \sigma_y\,[1 + \ln(1 + r/\rho)] \quad [1]$$

from slip-line field theory, where σ_y is the uniaxial yield stress in tension and r is the distance from the notch tip: and

$$r_c \approx 0.12\,(K_A/\sigma_y)^2 \quad [2]$$

from linear elastic theory, where r_c is the radius of

Table III. Ambient Temperature Instrumented Charpy Results for Alloy AISI 4340 in the Oil Quenched Condition, Showing the Variation of Charpy Impact Energy and Apparent Fracture Toughness (K_A) with Notch Root Radius (ρ).

Austenitizing Temperature, °C	Notch Root Radius, ρ		Charpy Impact Energy		Apparent Dynamic Fracture Toughness, K_A	
	in.	μm	ft lb	J	ksi$\sqrt{in}$	MPa$\sqrt{m}$
1200-870	0	0	1.5	2.03	48.8	53.7
	0	0	2.4	3.25	54.2	59.6
	0.0011	27.9	2.3	3.12	46.4	51.0
	0.0038	96.5	2.3	3.12	45.3	49.8
	0.0094	238.8	2.0	2.71	48.7	53.5
	0.0094	238.8	6.0	8.13	54.8	60.2
	0.0115	292.1	5.0	6.78	52.9	58.1
	0.0171	434.3	3.0	4.07	62.0	68.1
	0.0230	584.2	3.0	4.07	69.2	76.1
870	0	0	1.0	1.36	30.0	33.0
	0	0	1.0	1.36	33.3	36.6
	0	0	1.0	1.36	36.6	40.3
	0.0012	30.5	2.3	3.12	48.7	53.5
	0.0027	68.6	3.0	4.07	63.4	69.7
	0.0033	83.8	4.5	6.10	78.2	85.9
	0.0084	213.4	5.7	7.73	95.5	105.0
	0.0100	254.0	7.5	10.2	105.7	116.2
	0.0133	337.8	9.5	12.9	126.4	138.9
	0.0175	444.5	12.3	16.7	132.5	145.6
	0.0220	558.8	13.5	18.3	137.1	150.7

the critical plastic zone size, then at failure, when $\sigma_{yy}^{max} = \sigma_F$, and $r = r_c$,

$$K_A \approx 2.9\,\sigma_y\,[\exp(\sigma_F/\sigma_y - 1) - 1]^{1/2}\rho^{1/2}. \qquad [3]$$

Eq. [3] shows that the apparent fracture toughness (K_A) varies with $\rho^{1/2}$, as found experimentally and shown in Fig. 2. However, for some radius ($\rho < \rho_0$), K_A is independent of ρ and has the same value as K_{Ic}, measured in standard ASTM specimens[5] which have been fatigue precracked before testing. Consequently, Tetelman and coworkers[9,10] proposed that

$$K_{Ic}\ (\text{or } K_{Id}) \approx 2.9\,\sigma_y\,[\exp(\sigma_F/\sigma_y - 1) - 1]^{1/2}\rho_0^{1/2}. \qquad [4]$$

The parameter ρ_0, the "effective" or limiting root radius, is a measure of the extent of the process zone ahead of the crack (the "characteristic distance") over which the critical stress σ_F must exist to cause failure, and is related to the microstructural feature which controls fracture (such as slip or twin band spacings, grain size or inclusion spacing).[13-15] The characteristic distance (l) represents the minimum distance from the notch where the critical fracture event can occur (*i.e.*, where $\sigma_{yy} > \sigma_F$), and is only important where the maximum tensile stress (σ_{yy}^{max}) is very close to the notch tip.[14,15] This is the case ahead of sharp cracks,[16] where σ_{yy}^{max} at fracture is located at a distance ahead of the crack tip which is generally *smaller* than the characteristic distance. Ahead of

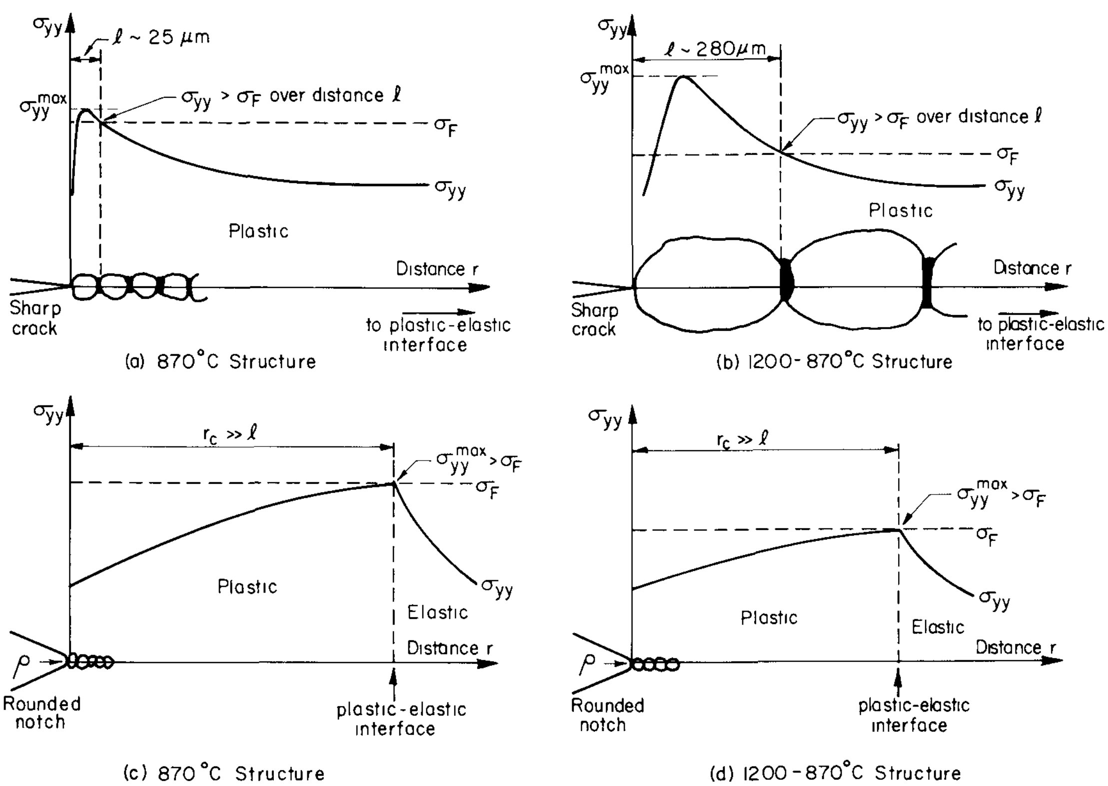


Fig. 3—Schematic representation of the distribution of tensile stress (σ_{yy}) at distance (r) ahead of stress concentrator at failure for (a) 870°C structure with sharp crack ($\rho < \rho_0$), (b) 1200-870°C with sharp crack, (c) 870°C structure with rounded notch ($\rho > \rho_0$), (d) 1200-870°C structure with rounded notch. Critical fracture event occurs when $\sigma_{yy} > \sigma_F$ *over* characteristic distance (l) ahead of sharp crack, or when $\sigma_{yy}^{max} \geq \sigma_F$ *at* the plastic-elastic interface ($r_c \gg l$) ahead of rounded notch. Toughness of 1200-870°C structure is greater ahead of sharp crack because characteristic distance (l) is larger, toughness of 870°C structure is greater ahead of rounded notch because fracture stress (σ_F) is larger.

rounded notches ($\rho > \rho_0$), however, σ_{yy}^{max} is located at, or just behind,[17] the plastic-elastic interface, and thus at failure the critical fracture event is occurring at a distance from the crack tip which is *large* compared with the characteristic distance.* A schematic representation of these fracture events is shown in Fig. 3.

The application of this model to the present results is complicated by the fact that it is only strictly valid for stress-controlled fracture, and was derived from data on cleavage in mild steel.[9] The failure mechanisms in 4340 steel are far more complex (see Fig. 1), but at least the dominant fracture mechanism for both structures can be considered to be largely stress-controlled (*i.e.* quasicleavage and intergranular cracking). In addition, the general form of the relationship between toughness and root radius shown in Fig. 2, and the existence of a limiting root radius would appear to be unchanged even for strain-controlled ductile fracture.[18]

Examination of Fig. 2 in the light of the above theory reveals i) the critical fracture stress, σ_F, is *smaller* in the 1200-870°C structure compared to the 870°C structure, as indicated by the differing slopes of the curves (see Eq. [3]). ii) The limiting root radius, ρ_0, is *larger* for the 1200-870°C structure compared to the 870°C structure. iii) For both structures, the value of ρ_0 is of the same order as the prior austenite grain size. These observations indicate that the higher austenitizing treatment has caused a reduction in critical fracture stress, but increased the critical distance over which it must be exceeded for failure. The increase in characteristic distance appears to be associated with the larger grain size. The decrease in σ_F is probably the result of grain boundary embrittlement due to segregation of impurity elements to the smaller grain boundary surface area of the larger grain-sized material-typically, segregation of phosphorous leading to intergranular cracking,[19,20] and segregation and precipitation of sulfur leading to intergranular fibrous rupture.[21,22]

This is consistent with the fractographic evidence in Fig. 1 which indicates that some form of grain boundary embrittlement, during the higher temperature austenitizing treatment, has lead to a change from predominately quasicleavage failure, seen in the 870°C structure, to intergranular failure in the 1200-870°C structure. The segregation of phosphorous to austenite grain boundaries would be more effective in causing embrittlement in the 1200-870°C structure because of the considerably larger grain size, resulting in intergranular cracking (Fig. 1(*b*)). The intergranular fibrous rupture, also observed in the 1200-870°C structure, would occur from the dissolution of manganese (and possibly chromium) sulfide inclusions at 1200°C, which then precipitate at lower temperatures as a fine network of sulfides on prior austenite grain boundaries[21,22] (see Fig. 1(*c*)). The 870°C structure shows no such evidence of intergranular failure (Fig. 1(*a*)). It is then suggested that this grain boundary embrittlement is responsible for the lower fracture stress, σ_F, of the 1200-870°C structure.

Thus, in summary, K_{Id} (and K_{Ic}) will be increased in the larger grained 1200-870°C structure because the fracture stress ahead of the sharp crack must be exceeded over a much larger distance (Fig. 3(*a*) and 3(*b*)). In Charpy tests, however, where the larger root radius causes the critical fracture event to occur much further away from the notch tip, the lower fracture stress of the 1200-870°C structure leads to an inferior K_A and thus lower impact energy with respect to the 870°C condition (Fig. 3(*c*) and 3(*d*)).

*A fuller explanation of how the stress distribution ahead of sharp and rounded notches can influence fracture behavior can be found in Ref. 14 and 15.

DISCUSSION

It can be seen from this work that the explanation for the differences in toughness, as evaluated by fracture toughness and Charpy tests, may be attributed to the differing response of each structure to the influence of notch root radius on toughness. An explanation of this effect in terms of microstructure is more complex. It has been suggested that the improvement in K_{Ic} from high temperature austenitizing can be attributed to a) dissolution of carbides at the high solution temperatures;[1,23] b) retention of austenite films around martensite plates;[1,23] c) elimination of proeutectoid ferrite[24] and upper bainite[23-25] due to increased hardenability from the larger grain size; or d) lack of twinned martensite plates in the as-quenched structure.[26,27] Such microstructural modifications, however, should also lead to increased impact energies, *which are not observed*. The present analysis indicates that the increase in K_{Ic} (and K_{Id}), following high temperature austenitizing, results merely from a significant increase in effective root radius ρ_0.

The magnitude of ρ_0 should relate to the particle spacing or grain size, depending upon the mechanism of failure. For particle-controlled fracture (*e.g*, fully fibrous rupture) the increase in ρ_0 could be attributed to increased carbide spacings, although, in the present case, these particle spacings do not relate numerically to the values of ρ_0 obtained. For cleavage and intergranular fracture, on the other hand, it has been shown[15,28] that the magnitude of ρ_0 is of the order of the grain size (*i.e.* approximately equal to two grain diameters for cleavage in mild steel,[15] and 1 to 1 1/2 grain diameters for intergranular failure in low alloy, En30A steel).[28] Since in the present results, the principal fracture modes are either cleavage or intergranular cracking, and the numerical values of ρ_0 are of the same order as the grain diameters, it must be concluded (although tentatively) that the increase in ρ_0, and hence K_{Ic}, for the high temperature austenitizing treatment, is a result of the order of magnitude increase in grain size.

It is clear that the use of high temperature austenitizing, instead of conventional heat-treatment, can result in steels of much improved fracture toughness (K_{Ic}), but the accompanying low Charpy impact energies must limit the practical usefulness of such procedures. Other results[1-4,29] in the literature substantiate this claim.* For example, it has been reported[29] that an Fe/3.85 pct Mo/0.18 pct C steel has a K_{Ic} = 93 ksi$\sqrt{\text{in.}}$ (103 MPa $\sqrt{\text{m}}$) after oil quenching from 1200°C,

*AISI 4130 and 4330 steels appear to be exceptions, although even these steels, when tempered above 300°C, show superior Charpy impact energies after austenitizing at 870°C, compared to the high temperature austenitized treatments.[3]

and yet the Charpy impact energy for this condition is a mere 11 ft lb (15 J). It should also be noted that the tensile ductility properties of low alloy steels austenitized at high temperatures can be greatly reduced. This can be seen from results[3] obtained for AISI 4130, 4330, 4140, 4340, 300M and D6-AC steels in both as-quenched and quenched and tempered conditions. In the present results, for instance, the reduction in area for as-quenched 4340 is only 8 pct in the 1200-870°C structure, compared to 31 pct in the 870°C structure (Table I). The percentage elongation is similarly reduced (by a factor of 3), and this is consistent with a lower "true" fracture stress in the 1200-870°C structure. Furthermore, the considerably larger prior austenite grain size of steels austenitized at higher temperature can lead to increased dangers from temper embrittlement[19] and tempered martensite embrittlement,[30] if the steel is subsequently tempered. The smaller grain boundary area associated with large grain sizes implies that segregation of impurity elements, should it occur, will be more effective in producing subsequent embrittlement because the relative degree of grain boundary coverage produced by a given amount of solute can be correspondingly greater.[28] Thus, although there is evidence,[31,32] in certain molybdenum-containing secondary hardening steels, that austenitizing at 1200°C can be beneficial to strength as well as fracture toughness (no Charpy impact values were reported), it is felt that, in general, the use of high temperature austenitizing treatments for ultrahigh strength, low alloy steels, in place of conventional low temperature treatments, should be regarded with extreme caution.

Finally, it is considered that the phenomenom described in this paper has an important bearing on what procedures should be carried out in alloy design to evaluate material toughness. Although K_{Ic} (and K_{Id}) measurements are obviously *necessary*, it appears no longer *sufficient* to grade toughness solely in terms of these parameters, since they only indicate material resistance to fracture ahead of sharp cracks. For the present steel, for example, grading material toughness by K_{Ic} (or K_{Id}) alone would lead to the conclusion that the steel austenitized at 1200°C was tougher. Yet, if in service the steel contained a blunt indentation, the structure resulting from the conventional 870°C heat-treatment would show superior resistance to fracture for all notch root radii greater than 0.002 in. (0.05 mm). Thus, it is suggested that the toughness evaluation of a material must include a measure of resistance to fracture ahead of both sharp and rounded notches. It appears insufficient to grade material toughness solely on the basis of K_{Ic} *or* Charpy impact energy values alone. An assessment from both sharp crack fracture toughness and rounded notch impact energy testing is required.

CONCLUSIONS

1) It has been verified that a high temperature austenitizing treatment of 1200°C (followed by a step quench to 870°C) improves the fracture toughness of as-quenched AISI 4340, but gives rise to inferior Charpy impact energies compared to conventional 870°C austenitizing.

2) This effect was shown to be independent of shear lip energy differences, and to be independent of variation in strain rate imposed by K_{Ic} and impact tests.

3) The effect was found to result from the differing behavior of each structure to the influence of notch root radius on toughness.

4) The structure obtained after austenitizing at 1200°C was found to show superior toughness for failure ahead of sharp cracks (*i.e.* in K_{Ic} and K_{Id} tests) due to its larger "effective" root radius for fracture. This was tentatively associated with the larger prior austenite grain size.

5) The structure obtained after conventional austenitizing at 870°C was found to show superior toughness for fracture ahead of rounded notches (*i.e.* in Charpy impact tests) due to its larger critical fracture stress for failure.

6) The use of high temperature austenitizing treatments, in place of conventional heat-treatments, for ultrahigh strength, low alloy steels is questioned, because the structures which result appear to show low Charpy impact energies and much lower ductility, and may give rise to increased susceptibility to embrittlement during tempering.

7) It is suggested that evaluation of material toughness must include a measure of resistance to fracture ahead of both sharp and rounded notches. It appears insufficient to grade material toughness solely on the basis of K_{Ic} *or* Charpy impact energy values alone. Both sharp crack K_{Ic} *and* rounded notch impact energy values are required.

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REFERENCES

1 G. Y. Lai, W. E. Wood, R. A. Clark, V. F. Zackay, and E R Parker *Met. Trans*, 1974, vol 5, p 1663.
2 D Dulieu Discussion to paper by V F Zackay, E. R. Parker, and W. E. Wood, *Proc Third Intl Conf. on the Strength of Metals and Alloys,* vol. 2, p. 383, Institute of Metals/Iron and Steel Institute, Cambridge, England, Aug. 1973
3 W. E. Wood, E. R. Parker, and V. F. Zackay Rept. No. LBL-1474, Lawrence Berkeley Laboratory, University of California, Berkeley, Calif. 94720, May, 1973.
4 A A. Anctil and E B Kula unpublished research, Army Materials and Mechanics Research Center, Watertown, Mass., 1975.
5 Annual Book of ASTM Standards (American Society for Testing and Materials), ASTM Standards E399-74, Philadelphia, 1974.
6. Annual Book of ASTM Standards, (American Society for Testing and Materials), p. 276, ASTM Standards E23-72, Philadelphia, 1972.
7 W. L. Server and A. S. Tetelman *Eng Fract Mech.*, 1972, vol 4, p. 367.

8. W. L. Server, D. R. Ireland, and R. A. Wullaert Dynatup Technical Rept. No TR 74-29R, Effects Technology, Inc., Santa Barbara, California 93105, November 1974.
9. T. R Wilshaw, C. A Rau, and A. S. Tetelman *Eng. Fract. Mech*, 1968, vol. 1, p. 191.
10. J. Malkin and A S. Tetelman· *ibid,* 1971, vol. 3, p. 151.
11. P. T Heald, G M. Spink, and P. J. Worthington *Mater Sci Eng.*, 1972, vol 10, p. 129.
12. S. Ensha and A. S. Tetelman. UCLA Tech. Rept , No. 14, UCLA-ENG-7435, School of Engineering and Applied Science, University of California, Los Angeles, Calif 90024, May 1974
13. A. H. Cottrell *The Mechanical Properties of Matter,* p 365, John Wiley & Sons, New York, 1964.
14. A. S. Tetelman, T. R Wilshaw, and C. A. Rau *Int J Fract Mech*, 1968, vol 4, p. 147.
15. R O Ritchie, J. F. Knott, and J. R Rice. *J Mech Phys Solids,* 1973, vol 21, p 395.
16. J. R. Rice and M. A. Johnson *Inelastic Behavior of Solids,* M. F Kanninen *et al,* eds , p. 641, McGraw Hill, New York, 1970.
17. J. R. Griffiths and D. R J. Owen *J. Mech Phys Solids,* 1971, vol 19, p 149.
18. C G. Chipperfield and J. F Knott *Metals Technol*, 1975, vol. 2, p. 45.
19. B J Schulz and C. J McMahon, Jr. ASTM STP No. 499, p. 104, American Society for Testing and Materials, 1972
20. G Clark, R O Ritchie, and J F Knott *Nature Phys. Sci*, 1972, vol 239, p. 104.
21 B. J Schulz and C J. McMahon, Jr · *Met Trans*, 1973, vol 4, p 2485.
22 R O Ritchie and J. F Knott *ibid,* 1974, vol. 5, p. 782
23. G. Thomas *Iron Steel International,* 1973, vol 46, p 451
24. G Y Lai *Mater Sci. Eng.*, 1975, vol 19, p 153
25. W E Wood *Eng. Fract. Mech*, 1975, vol 7 p 219
26. S K Das and G Thomas *Trans ASM,* 1969, vol 62, p 659,
27 J A McMahon and G Thomas *Proc Third Intl Conf. on the Strength of Metals and Alloys,* vol 1, p 180, Institute of Metals/Iron and Steel Institute, Cambridge, England, Aug. 1973.
28. R O Ritchie, L C E Geniets, and J F Knott· *ibid,* p 124.
29 R A. Clark and G Thomas *Met Trans. A,* 1975, vol 6A, p 969
30. J R Rellick and C J McMahon, Jr · *Met Trans*, 1974, vol 5, p. 2439
31 V F. Zackay, E R Parker, R D Goolsby, and W. E Wood· *Nature Phys. Sci*, 1972, vol 236, p 108.
32 T Tom D Eng. Thesis, Rept. No LBL-1856, Lawrence Berkeley Laboratory, University of California, Berkeley, Calif 94720, Sept 1973