CORRELATIONS BETWEEN FRACTURE TOUGHNESS
AND MICROSTRUCTURE IN 4140 STEEL.

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By

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ABSTRACT

Correlations between the microstructure of an ultra-high strength steel and material resistance to fracture, as measured by blunt notch Charpy impact tests and "sharp crack" K_{IC} tests, were investigated for a standard 870°C/oil and an experimental 1175°C/oil austenitizing treatment. The increase in "sharp crack" toughness with higher temperature austenitizing treatments, for the as-quenched and 200°C/oil temper conditions, was rationalized by a fracture criterion based on the notion that for fracture to occur, a critical strain, ϵ_f^* , must be achieved over some critical distance, δ^* . The lath colonies were identified as the fracture controlling microstructural unit, and hence, their size was considered to be the critical distance, δ^* . Toughness in the 300°C/1 hour and 400°C/1 hour temper conditions, for which the mechanical data indicated an embrittlement, was clearly controlled by the cementite morphology in conjunction with the prior austenite grain size. Attempts to rationalize toughness in these temper conditions, using a stress-controlled fracture criterion, were unsuccessful and led to physically unreasonable results. In the 500°C/1 hour temper condition, stable crack growth and periodic ridge patterns were observed. Fracture toughness differences between the 870°C and 1175°C austenitizing treatments were qualitatively rationalized by the nature of the respective fracture morphologies. Good correspondence between J_{IC} and the so-called tearing modulus, T , as indicators of "sharp crack" fracture toughness, was observed.

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1. INTRODUCTION

1.1 Scope of the Research

The purpose of this research program was to investigate correlations between the complex microstructure of a typical ultra-high strength quenched and tempered steel, namely 4140, and material resistance to fracture. Specifically, the effect of such microstructural parameters as prior austenite grain size, lath, and lath colony size, and carbide morphology on strength, ductility, Charpy V-notch impact energy, plane strain fracture toughness, and stable crack growth were examined. The results are presented in section 3. In view of the many recent investigations into the use of high austenitizing temperatures to improve sharp crack fracture toughness (along with the seemingly contradictory results as determined by blunt notch Charpy tests), particular attention was focused on material response to a conventional 870°C/oil austenitizing treatment versus an 1175°C/oil treatment, for various stages of temper.

In attempting to establish a fundamental understanding of the micromechanics of the fracture processes in this class of steels, and indeed, in other materials as well, it seems clear that an important first step is the identification of those microstructural features which actually control the fracture process, and hence, the fracture toughness. Such an understanding could serve as a rational guide in the microstructural design of these alloys for optimal strength-toughness properties. For example, microvoids which initiate at carbide-matrix interfaces, and grow and coalesce with plastic deforma-

tion, provide the actual mechanistics of at least the inception of "ductile" fracture in these steels. Since carbides often precipitate preferentially at interfaces, it must be anticipated that prior austenite grain boundaires, lath, and lath colony boundaries could well determine the carbide spacing and ultimately provide "easy paths" for void coalescence and crack advance. Hence the sizes of these microstructural features were measured and compared with what appeared to be important morphological features on the fracture surface. The morphology of main crack profiles (in longitudinally sectioned CT specimens) and subsurface secondary cracks (in both CT specimens and Charpy bars) were also examined, with the intent of determining just what features seem to control the propagation of cracks, and hence the resistance of the material to fracture. With this sort of knowledge, one can begin to postulate micromechanically based fracture criteria with which, along with continuum analyses of crack tip stress and strain fields, fracture toughness can be predicted or at least rationalized.

As an example, in the case of specimens tempered in the 300°C to 400°C embrittlement range, it was clear that the carbide (cementite) morphology, in conjunction with the prior austenite grain size, controlled the fracture process. For specimens in the as-quenched and 200°C/1 hour temper conditions (for both austenitizing treatments), although not as obvious as for the embrittled specimens, it appeared that lath colony size was very influential in determining toughness.

In this latter case, a fracture criterion which is useful in rationalizing toughness differences between specimens austenitized at 870°C and 1175°C is based on the notion that for fracture to occur, a critical strain, ϵ_f^* , must be achieved over some suitable distance, δ^* . ϵ_f^* is typically fixed by the scale and character of the microstructure, i.e. by the carbide structure, whereas δ^* is a measure of the mean spacing of the fracture initiating sites. In the present case, δ^* would appear to be related to the lath colony size.

Criteria of this sort can be of considerable aid in evaluating the toughness of similar microstructures in other materials, and along with a fundamental understanding of the micromechanics of the fracture processes involved, facilitate the design and development of new tougher ones.

1.2 Background

During the past four decades, an extensive literature concerning the correlations between the mechanical properties of high strength, quenched and tempered steels and their complex microstructure has evolved. In the past 30 years or so, and particularly since the advent of the stress intensity approach to linear elastic fracture mechanics [12], an increasing emphasis has been placed on the property of toughness. This vast body of information offers some very general guidelines for microstructural control which usually suggest that refined microstructures provide optimum toughness-strength properties. Refined particle substructures, for example, have been shown to increase toughness [13], due to the now well understood effect of large particles (notably carbides and sulfides in steels) in initiating voids or micro-cleavage cracks. Platelike carbide morphologies, formed at grain, lath, and lath colony interfaces during tempering of lower bainite or martensite in the 250-400°C range, or by precipitation at the lath boundaries in upper bainitic microstructures, are also generally known to be deleterious to toughness (especially in conjunction with alloy element segregation to these interfaces). There are numerous reports, for example, of superior toughness properties (at comparable strength levels) in untempered lower bainitic structures (with a fine, relatively homogeneous carbide

distribution) as compared with upper bainitic or tempered martensitic structures. However, despite the long standing validity of the above generalities, there have been more than occasional observations that, by forming coarse austenitic grains and thus coarse martensite and bainite lath colonies, an increased fracture toughness can be achieved (at least for as quenched and lightly tempered structures). Such coarser structures are readily formed at austenitizing temperatures in the 1100 - 1200°C range. This increase in toughness, though, is limited to measures of resistance to fracture initiation at initially sharp cracks. More recently these sort of observations have been discussed by Parker and Zackay and co-workers [15,16], Wood [17,18], and Ritchie and co-workers [19,20,21]. The work in these latter studies focused on a number of (ultra) high strength steels, including a 4140 alloy but with, by far, the major discussion concerned with 4340 steel.

Essentially, the results of these studies suggested that when 4340 steels were austenitized at temperatures near 1200°C, versus the more standard commercial 870°C, the plane strain "sharp crack" fracture toughness was increased by a factor of approximately 1.5. Other measures of toughness, such as room temperature Charpy impact energy and work to fracture in a tension test, as well as ductility measures (reduction in area and true strain to fracture), were

decreased. Wood [18] emphasized some results for as-quenched microstructures which did not show decrements in Charpy toughness, but as pointed out by Ritchie et al. [21], or in any event as is already well documented in the early literature, the trend in this class of steels is for coarse microstructures to lead to lower ductilities and reduced impact energies.

Ritchie and co-workers [20] offered the beginnings of an explanation for the apparent inconsistency in the evaluation of toughness in these steels, by discussing the fracture process in terms of the nature of the stress-strain field ahead of an initially sharp crack versus a blunt notch. In short, their premise is as follows: if a fracture criterion is adopted which requires that a critically large stress must be achieved over a critically large enough distance (to incorporate crack initiating particles preferentially situated on interfaces, for example), then coarse microstructures can be more resistant to fracture at sharp cracks. This is due to the relatively narrow stress distribution ahead of a sharp (or even initially sharp, but blunted) crack under conditions of small scale yielding in plane strain. This seems to be a useful approach, as will be discussed in later sections, although modifications in detail are required if results which are qualitatively consistent with experiment are to follow.

For the 4140 steel investigated in the current work, we have first attempted to identify the "microstructural unit" (a very

appropriate phrase borrowed from the work of Bernstein and co-workers on cleavage fracture in steel)[22] which actually controls the fracture process. We have then applied a "strain-controlled" fracture criterion by using the results of Rice and co-workers [23] on crack tip stress-strain fields. As discussed in Section 4, this leads to a reasonable description of the observations on the as-quenched and 200°C/1 hour temper conditions, i.e. those tempers where an increase in toughness would be of the most practical importance. Discussion of the fracture mechanisms in the other temper conditions is also presented in Section 4. Finally, in Section 5, we describe what we feel are important extensions of this sort of work, and what direction future research on this problem should take.

2. EXPERIMENTAL PROCEDURE

2.1 Heat Treatment

Experiments were carried out on an Al-Si killed AISI 4140; the chemical analysis, as provided by the Bethlehem Steel Corporation; is found in Table 2.1. The ingots had been mill homogenized and hot rolled. Mechanical test specimens were austenitized in an Argon atmosphere for 1 hour at $870^{\circ}\text{C} \pm 10^{\circ}\text{C}$ or $1175^{\circ}\text{C} \pm 15^{\circ}\text{C}$, and directly quenched into well agitated room temperature oil. Blanks for optical metallographic examination ($6.4 \times 6.4 \times 12.7$ mm) were, in addition, austenitized at $1000^{\circ}\text{C} \pm 10^{\circ}\text{C}$ or $1100^{\circ}\text{C} \pm 10^{\circ}\text{C}$. Subsequent to the austenitization treatment, specimens were tempered for 1 hour at 200°C , 300°C , 400°C , and 500°C , and quenched into oil. Mechanical test specimens were heat treated in a slightly oversize (2.54 mm for tensile bars, 1.52 mm for Charpy bars, and 1.27 mm for compact tension specimens) configuration to avoid surface decarburization effects.

2.2 Microscopy

Sections for optical microscopic examination were obtained from the blanks and from various mechanical test specimens, as designated. These sections were cut (flood cooling was employed to avoid tempering) and polished using standard preparation techniques. In the final stage (.05 micron alumina solution), a polish-etch-polish-etch procedure was employed to eliminate possible false structures and polishing induced surface deformations. Three percent Nital and 4 percent Picral solutions, either separately or in combination, were employed as etchants. A third etchant, comprising 1 gram of sodium tridecylbenzene sulfonate dissolved in 100 ml of saturated aqueous picric acid[1], provided excellent grain boundary delineation. This particular etchant should be used when hot (just below

the boiling point) and specimens should be given a pre-temper of approximately 350°C for 1 hour.

Relevant microstructural features were characterized using methods of quantitative microscopy [2]. In determining the inclusion volume fraction, point counts were obtained on 35 fields of view for each temper at the 870°C and 1175°C austenitizing treatments. An average of 10 fields (and as many as 15 at critical points), at each of the austenitizing temperatures, was counted to determine mean lath colony (packet) size and prior austenite grain size. Approximately 10 fields (obtained from replica transmission electron micrographs) at the high and low austenitizing treatments were counted to estimate mean lath size.

Replicas for transmission electron microscopic examination were obtained from polished sections cut from the hardness determination specimens. They were prepared using a standard two stage technique [3]. Carbon deposition (normal to the replica surface) and chromium shadowing (.003 gr at a 30 degree angle from the horizontal) were carried out under pressures of no more than $6.7 \times 10^{-3} \text{ N/m}^2$ ($5 \times 10^{-5} \text{ Torr}$). The replicas were viewed using a Jeolco JEM-7 operated at 80 kv accelerating potential.

All fracture surfaces were examined using scanning electron microscopy (AMR-1000A, operated at 20 kv accelerating potential). Specimens were sprayed with a protective coating (Krylon Crystal Clear) immediately after separation. This coating was readily dissolved in acetone (using an ultrasonic cleaner) just prior to examination. An EDAX (energy dispersive analysis of X rays) unit was used in conjunction with the SEM to provide an elemental analysis of the inclusions.

Subsurface microcracking was examined in a number of the fractured Charpy bars. The separated halves were first electroplated with Cu and Ni (to preserve the edges during polishing), and then sectioned longitudinally at the midplane using a precision wafering machine (with flood cooling). Fully opened crack profiles in the compact tension specimens were obtained using a displacement preserving technique wherein diagonally cut cylindrical pins were wedged into a cylindrical hole machined into the specimen (see Figure 2.1). Subsequent to unloading and removal from the test machine, the cracks were impregnated with epoxy and the specimens sectioned longitudinally at the midplane. Subsurface cracking was also viewed in the separated compact tension specimens following the procedure described above for the Charpy bars.

2.3 Mechanical Testing

Room temperature uniaxial tensile properties were determined using round button-grip bars (107.95 mm in length, reduced section is 38.1 mm in length and 5.08 mm in diameter) oriented longitudinally with respect to the rolled plate. The specimens were pulled in an Instron testing machine at a crosshead speed of 1.27 mm/sec initially and at 0.51 mm/sec subsequent to yielding. Post necking diameter measurements were obtained using a Riehle spring loaded lateral extensometer, and an extrapolated value of the area at fracture was determined for each temper.

Hardness measurements were taken on blanks (initially 19 x 16 x 12.7 mm) which had been sectioned to the mid-plane, ground flat, and polished to the 600 grit soft stage (15 micron silicon carbide paper; Beuhler Ltd.).

Dynamic toughness was measured using a Charpy V-notch test. Standard size specimens [4] were broken in a pendulum type impact machine at a hammer velocity of 3.4×10^3 mm/sec. Bars were tested in the L-T orientation (notation is in accordance with the system set forth in ASTM E399-72 [5]), with a second series of tests for the 870°C treatment conducted on bars in the L-S orientation. All Charpy tests were performed at room temperature.

Plane strain, quasi-static, sharp crack fracture toughness values were determined using 1T (11.7 mm thick due to directional constraints of the received stock) compact tension specimens in the L-T orientation. The specimen configuration was slightly modified to allow for load-line displacement measurements. [6] Specimens were fatigue precracked in tension-tension, on an Instron closed loop hydraulic test machine, using a sine-squared wave at 35 Hertz as the control function. K_{fatigue} levels were maintained at less than $0.6 K_Q (= K_{IC})$ in accordance with standards. Cracks were extended to a/w ratios ranging from 0.7 to 0.8. Since this is considerably beyond the range of validity of the conventionally employed compliance coefficient (see ASTM E399-72 [5]), the polynomial expression proposed by Srawley [7] (valid for $.2 < a/w < 1$) was utilized in the K_{IC} computations.

In initial tests, a thumbnail crack front developed during the loading, along with the formation of extensive shear lips at separation. To avoid this, all subsequent specimens were machined with 0.51 mm deep side grooves. Crack fronts in these specimens were straight (generally with less than 1 percent variation from edge to edge), and no shear lips were present. Initial crack lengths were determined by measuring

the fractured halves using a micrometer-stage equipped stereoscope, in accordance with ASTM E399-72 [5]. ASTM procedures were followed in checking the validity of the tests, and where valid, in computing K_{IC} .

For those cases where a valid K_{IC} test could not be obtained using the 12.7 mm thick specimens, a J resistance curve was calculated from the load-deflection data and J_{IC} determined. In this work, the single specimen J_{IC} determination test procedure developed by Clarke, Paris, et al. [8] was followed. An extremely sensitive differential compliance technique (developed by Hermann, Paris, et al. [8,9,10]) was employed in determining crack length changes. During the unloading and reloading portion of the test, a differential displacement signal, $d\delta$, is obtained by electronically subtracting a load-proportional signal, $\delta_{ELASTIC}$, from the load-line displacement signal, δ_{ACTUAL} . The constant of proportionality of the subtracting signal is the initial elastic compliance, $C_{ELASTIC}$ with the fatigue pre-crack fully opened. The signal, $d\delta$, when amplified and graphed versus load, will remain vertical until an increment of crack growth (more precisely, until tip blunting and plasticity preceding growth) has occurred. The slopes that are recorded once the crack has grown are the inverse of the compliance change (differential compliance), dC , with respect to the initial elastic compliance. By recording differential compliance, instead of compliance, a more sensitive scale can be employed, enabling the detection of crack length changes on the order of 0.1 mm. At the start of the test, the elastic compliance signal is scaled so that a vertical line is produced (e.g. graphed on the x-y plotter) when the specimen is loaded in the elastic region (at loads greater than the crack opening load). $C_{ELASTIC}$

is then given by [10]:

$$C_{ELASTIC} = \frac{S_u \times V_p}{S_p \times V_u \times R} \quad (2.1)$$

S_u = the displacement scale

S_p = the load scale

V_u = the strain potentiometer reading

V_p = the load potentiometer reading, and

R = the ratio of displacement signal sensitivity to load signal sensitivity.

A schematic diagram of the electronic system is provided in Figure 2.2.

The relations used in calculating the initial crack length and crack length changes are found in Appendix A.

In order for the scheme described herein to be fully effective, special precautions must be taken to minimize friction effects which will be manifested as a hysteresis in the loading and unloading curves. Bearings are incorporated into the hangar plates at the point of contact with the loading pins, to accommodate rotations accompanying bending of the specimen (for deeply cracked CT specimens with an $a/w > 0.6$). However, under load, the pins will bend and jam against the bearings. By machining a necked down section in the hangar plates, shown in Figure 2.1, elastic flexing is permitted, and jamming is effectively prevented. This freedom of movement results in nearly hysteresis-free curves, as shown in the example in Figure 2.3.

3. RESULTS

3.1 Microstructures

Optical metallographic examination revealed the gross features of the microstructures, as shown, for example, in Figure 3.1 for the as-quenched blanks austenitized at various temperatures. The microstructure is largely composed of laths, with what appear to be martensite plates occasionally observed along prior austenite grain boundaries (particularly evident in the specimens austenitized at temperatures in excess of 870°C--see Figure 3.1e as an example) and internally at random orientations with respect to the laths (Figure 3.1e). Pearlite is not present in any of the as-quenched specimens.

The prior austenite grain size was found to increase progressively with increasing austenitization temperature (see Figure 3.2), varying by a factor of 7, from an average of approximately 20 microns for specimens austenitized at 870°C, to 140 microns for specimens treated at 1175°C. The mean lath colony size (alternately referred to in the literature as the packet size) varied by a factor of two (see note below), from approximately 22 microns to 42 microns. The lath colony size obtained for the specimens austenitized at 870°C is most likely an overestimate, since there were many instances in which the laths were so randomly oriented that an unambiguous determination of a colony boundary was effectively precluded. The results of point counts obtained from these micrographs, therefore, undoubtedly suffer somewhat in objectivity. Indeed, as shown by Figure 3.1c, a typical field of view contains several grains with more than one lath orientation. This leads one to a lath colony size estimate of roughly 10 microns, assuming

a reasonable value of two different orientations per grain. In the specimens austenitized at temperatures greater than 870°C, the tendency for laths to group in well organized parallel arrays was more pronounced and point counts were correspondingly more accurate. As a further observation, laths seldom traversed an entire grain; cases where this did occur were limited to the smallest grains in the specimens austenitized at 870°C.

Optical microscopy revealed the presence of inclusions, the predominate type shown by EDAX analysis to be rich in manganese and sulfur (see Figure 3.3). Small, dark, irregularly-shaped particles located within many of the MnS inclusions were found to be calcium rich. Optical micrographs showing the MnS inclusions as viewed on the L-W plane and the L-T plane, are provided in Figures 3.4a and 3.4b, respectively. A calcium rich particle is clearly evident in the micrograph of Figure 3.4b. A low magnification view of the L-T plane, provided in Figure 3.5, shows the banding and inclusion segregation to lean zones which occurred during mill-rolling and homogenization, and was evident even after austenitization at 1175°C. Views of the W-T plane also showed this effect; on the L-W plane, ellipsoidal regions of varying etching intensity were observed. The volume fraction of the MnS inclusions, listed in Table 3.1 for each temper of the 870°C and 1175°C austenitizing treatment, averaged 0.11 percent. They ranged in size from 5 to 50 microns, with an average of approximately 10 microns. A second inclusion type, only occasionally observed in the heat of 4140 currently under investigation, was found previously (during a preliminary investigation of a similar heat of 4140) to be titanium-rich.

Transmission electron micrographs of replicas obtained from suitably polished and etched sections are presented in Figures 3.6 and 3.7 for the

870°C and 1175°C austenitizing treatments, respectively. In the specimen austenitized in the conventional manner, fairly coarse precipitates are quite clearly evident in virtually all the laths, whereas in the large-grained specimen, many of the laths appear to be precipitate-free.

Furthermore, the precipitates that are observed in this latter structure are noticeably finer than those of the 870°C treatment. Extensive observation of the as-quenched structures formed by both the 870°C and the 1175°C austenitizing treatments failed to reveal the presence of any lath boundary precipitate. Aside from the aforementioned size effect of the austenitizing treatment, no striking differences in the nature of the precipitates (which generally had a platelet morphology) were observed. On tempering at 200°C for 1 hour, the only discernable change in both the large and small-grained materials was a slight coarsening of the precipitates.

A temper of 300°C for 1 hour resulted in a considerable coarsening of the precipitates, in addition to the precipitation of long plate-like particles at the lath boundaries in the structure formed by the 870°C austenitizing treatment. Lath boundary precipitates, although present in the large-grained material after this temper, were not an obvious feature. In the fine-grained material, tempers of 400°C and 500°C had little apparent effect on the particle substructure as observed at 300°C. The overall carbide distribution was quite homogeneous. Tempering of the large-grained material at 400°C and 500°C, however, produced a significant coarsening and agglomeration of the carbides; lath boundary precipitates were readily apparent in these tempers. A heterogeneity, resulting in lath-shaped regions of uniform carbide distribution interspersed with

regions characterized by large particles oriented on what appear to be lath boundaries, had also developed.

Lath sizes for both austenitizing treatments varied over a wide range, thus making a quantitative characterization difficult. Estimates of lath length indicate a range of from 1.5 to 8 microns, with an average of approximately 3.5 microns, and a range of from 3 to 10 microns, with an average of roughly 6 microns, for the 870°C and 1175°C austenitizing treatment, respectively. Lath widths averaged 0.4 microns and 0.7 microns for the small and large-grained material, respectively.

In summary, the transmission electron microscopic examination revealed that in the 4140 steel currently under investigation, pronounced differences exist between the structures formed by the two austenitizing treatments. In the material austenitized at the conventional temperature, the as-quenched laths contain a uniform population of precipitates. This is in contrast to the material austenitized at 1175°C, where two distinct types of laths, one containing fine precipitates, and the other essentially precipitate-free, were observed. Furthermore, upon tempering the small grained material, long plates precipitated at the lath boundaries concurrent with a considerable coarsening of the precipitates located within the laths. A much less obvious lath boundary precipitation was observed in the large-grained specimens. With further tempering, i.e., at 400°C, a homogeneous distribution of precipitates developed in the small-grained material, in contrast to the heterogeneous plate-like carbide structure observed in the large-grained material.

3.2 Fractography

The results of the SEM examination of the compact tension specimens austenitized at 870°C and 1175°C are presented in Figures 3.8 to 3.10 and 3.11 to 3.14, respectively. All views are taken from the midpoint region of the specimens, just beyond the fatigue pre-crack. The fracture morphology of the fine-grained specimens, both in the as-quenched condition and in the 200°C/1 hour temper condition, is mixed transgranular ductile rupture and quasi-cleavage. Many of the cleavage facets are located within well-delineated regions most probably associated with either the lath colonies or the prior austenite grains. No obvious intergranular cracking was observed in either condition. A considerable amount of subsidiary cracking, resulting in the protrusion of large spikes from (or, alternately, of large incisions in) the fracture surface, was found in the as-quenched specimen. This sort of behavior was largely absent from the specimens given a 200°C/1 hour temper. Specimens tempered at 300°C for 1 hour also exhibited a mixed transgranular dimpled rupture and quasi-cleavage fracture morphology, although in this case, both the prevalence of the quasi-cleavage mode and the physical extent of many of the facets had increased. Isolated regions of intergranular cracking were also observed. Tempering at 400°C for 1 hour produced little, if any, discernible change in the fracture morphology from that observed in the 300°C/1 hour temper condition. No observations of the sort of subsidiary cracking behavior described for the as-quenched specimen were made in the specimens tempered at 300°C or 400°C. A temper of 500°C for 1 hour produced a fracture surface consisting of virtually 100 percent dimpled rupture, with little evidence of intergranular cracking or transgranular cleavage. Examination using a stereoscope at low

magnification revealed a regular, well-defined formation of ridges over the entire fracture surface. Similar ridges were also present in a small region (approximately 560 microns in width at the midpoint of the crack front) just at the head of the fatigue pre-crack in the specimen tempered at 400°C. A ridge region was just barely discernable in the 300°C/1 hour temper specimen.

In the large-grained material, both the as-quenched specimen and the specimen in the 200°C/1 hour temper condition exhibited a complex fracture morphology consisting of transgranular dimpled rupture, quasi-cleavage, and to a much lesser extent, intergranular cracking. The cleavage facets were generally found within well-defined regions (fringed by fine dimple networks) ranging in size from 30 to 80 microns. Only rarely did the size of a well-delineated cleaved region approach that of the prior austenite grain size. The sort of subsidiary cracking present in the fine-grained as-quenched specimen was completely absent in the corresponding temper condition of the large-grained specimens, and was only occasionally observed in the specimen tempered at 200°C. Tempering at 300°C produced a fracture morphology consisting primarily of intergranular cracking and dimpled rupture (both to approximately the same extent), with isolated regions of cleavage. A temper of 400°C resulted in a predominately intergranular mode of fracture, with regions of dimpled rupture interspersed amongst the grains. Specimens tempered at 500°C displayed an extremely fibrous intergranular cracking. Again, a regular, well-defined ridge pattern (covering the entire fracture surface beyond the fatigue pre-crack) was readily observed. Very small (<200 microns) regions just ahead of the fatigue pre-crack, appearing to have some ridge-like characteristics, were present in the specimens

tempered at 300°C and 400°C.

The fracture surfaces of the Charpy bars were also examined, and in all cases, fracture morphologies comparable to those described for the compact tension specimens were observed. At low magnifications, as viewed using a stereoscope, the fracture surfaces of the specimens in the as-quenched, 200°C/1 hour, and 300°C/1 hour temper conditions were characterized by typical fast fracture radial marks, with virtually no fibrous marks (here the macroscopic sense of fibrous is intended; on the microscopic level, the extent of ductile rupture in all these specimens was considerable). The specimen austenitized at 870°C and tempered at 400°C exhibited limited amounts of fibrosity; in the large-grained material of the same temper, the intergranular mode of cracking obliterated any other observations. The large and small-grained specimens in the 500°C/1 hour temper condition both were characterized by the same type of ridge pattern described previously. Shear lips, observed on all the separated Charpy bars, were found to increase in size with increasing absorbed impact energy. As a further observation, shear lips formed in specimens of the same temper (although at a different austenitizing temperature) were approximately equal in size.

Photomicrographs of subsidiary and main crack profiles are presented in Figure 3.15. Examination of a fractured, plated, and sectioned large-grained compact tension specimen revealed the presence of occasional trans-lath (transcolony) subsidiary cracking (Figure 3.15a) and the predominate intergranular mode of subsidiary cracking (see Figures 3.15b and c) which was typical of the as-quenched and 200°C tempers. The micrograph in Figure 3.15a shows quite clearly the change of crack direction as variously oriented

lath colonies are intersected. The fracture surface profile, shown in Figure 3.15c, also indicates this sort of apparent lath colony controlled fracture. Despite the large amounts of intergranular subsidiary cracking, smooth regions on the fracture surface itself (an example is shown in Figures 3.15b and c) are only occasionally observed. A sectioned Charpy bar (large-grained; 200°C/1 hour temper condition) exhibited wholly intergranular subsidiary cracking (shown in Figure 3.15d), but again, the fracture surface profile (Figure 3.15e) indicates a colony controlled behavior in the main crack.

A pinned and sectioned small-grained compact tension specimen, in the 200°C/1 hour temper condition, revealed a main crack profile that appeared to be predominately intergranular, with occasional transgranular regions. Examination of a small-grained, longitudinally sectioned Charpy bar in the 200°C/1 hour temper condition revealed a fracture surface profile indicative of colony, not grain boundary fracture. A micrograph of the main crack profile of a small-grained compact tension specimen tempered at 500°C for 1 hour, showing well-defined ridges with a wavelength of approximately 110 microns, is presented in Figure 3.15f.

The observations thus far indicate that in the as-quenched and 200°C temper conditions, lath colony size is the microstructural parameter controlling the actual fracture process, at least in the large-grained material. In the small-grained specimens, the lath colony size again appears to be the controlling parameter, although in this case the situation is not as readily determined (due to the aforementioned difficulties in obtaining accurate lath colony size measurements in the small-grained material). Subsidiary cracking, in most of the observed cases, involved

a process (namely intergranular cracking) which was different from the one seemingly controlling the propagation of the main crack. Subsequent to tempering at 300°C and 400°C for 1 hour, and concurrent with the observed embrittlement (see section 3.3), the fracture morphology indicated that microstructural features scaling to prior austenite grain size (whether the actual mode of failure is cleavage or intergranular cracking), were important. This is particularly clear in the large-grained specimens. In the 500°C/1 hour temper condition, a so-called zig-zag crack growth (resulting in the formation of ridges) directly involving neither the prior austenite grain size nor the lath colony size, was evident for both austenitizing treatments.

3.3 Mechanical Properties

The room temperature, uniaxial tensile properties are listed in Table 3.2. From these data, and as shown in Figure 3.16, it is evident that the true yield strength (0.2 percent offset) and the ultimate tensile strength (with load referred to the current area at necking) are not appreciably affected by the increase in austenitizing temperature. An average decrease of 15 percent in the true fracture strength is observed for specimens austenitized at 1175°C when compared to those specimens treated at the conventional temperature. Figure 3.17 presents a comparison of uniaxial tensile properties, as determined in the present investigation, with data obtained from the literature.

Hardness values, as shown by Fig. 3.16, were largely unaffected by the higher austenitizing temperature. Hardness readings, determined for specimens in the 600°C/1 hour and 700°C/1 hour temper conditions, indicate

that no secondary hardening occurs in our material, regardless of austenitizing treatment. Two ductility parameters, true strain to fracture(ϵ_f) and the reduction of area (R.A.), were measured. As shown in Figure 3.18, the specimens austenitized at 870°C exhibited considerably better ductility than those specimens austenitized at 1175°C. Furthermore, in the 300°C to 400°C temper range, the large-grained specimens displayed an embrittlement behavior which was largely absent in the fine-grained specimens.

Charpy V-notch impact energy, shown in Figure 3.19, followed a pattern similar to that of the ductility data. In this case however, embrittlement was observed for specimens treated at both austenitizing temperatures, albeit to a greater extent in the large-grained material. A series of specimens austenitized at 870°C, with the notch machined in the L-S orientation, yielded results virtually identical to the specimens in the same condition, but tested in the L-T orientation.

Plane strain, sharp crack fracture toughness (K_{IC}) values are listed in Table 3.3 (along with other pertinent information regarding the tests) and plotted in Figure 3.20. For tempers of 300°C or less, the specimens austenitized at 1175°C displayed, on the average, 30 percent greater toughness than the specimens treated at 870°C. With a temper of 400°C for 1 hour, the specimen austenitized at 870°C recovered the toughness lost due to embrittlement, whereas the large-grained specimens exhibited a further decrease in K_{IC} . Upon tempering at 500°C, the large-grained specimens showed full recovery, although they are less tough than similarly tempered small-grained specimens. Large and small-grained specimens in the 500°C/1 hour temper condition exhibited stable crack growth, as evidenced by smoothly varying load-load line displacement curves.

Figures 3.21 and 3.22 present the J versus crack growth (da) resistance curves for the $500^{\circ}\text{C}/1$ hour temper condition of the small and large-grained material, respectively. Various J_{IC} determination criteria are shown in the figures as solid and dashed lines intersecting the J versus da curve; application of these criteria yielded the data compiled in Table 3.4.

The value of J_{IC} utilized in the conversion to K_{IC} was determined by the equation $\Delta a = J/2 \sigma_{\text{flow}}$ (where $\sigma_{\text{flow}} = \frac{1}{2}(\sigma_{ys} + \sigma_{\text{ultimate}})$), shown in the figures as the solid line. The results of a validity check of each test are also provided in Table 3.4. Further discussion of the various J_{IC} determination criteria is provided in section 4.

The so-called dimensionless tearing modulus, $\left[T = \frac{E}{\sigma_{ys}^2} \frac{dJ}{da} \right]$, proposed by Paris et al. [39] is calculated for each J versus da curve and tabulated in Table 3.4. The slope ($\Delta J/\Delta a$) is determined along the linear portion of the curve and not, in this case, at the point of J_{IC} . It was generally observed that as J_{IC} increased, the value of the tearing modulus, T , also increased.

4 DISCUSSION

4.1 Toughness in the As-Quenched and 200°C/1 hour Temper Conditions

Recent efforts [19,20] to rationalize inconsistent toughness evaluations (by K_{IC} and Charpy V-notch impact tests) of 4340 austenitized at increased temperatures, have involved a consideration of the nature of the stress field ahead of a sharp crack versus a blunt notch, and the effects of the higher temperature austenitizing treatment on the so-called limiting notch root radius [24,25], ρ_0 . Briefly, for fracture processes considered to be "stress controlled," the increase in K_{IC} with high austenitizing temperatures was attributed to the coarsening of the fracture-controlling microstructural unit (i.e., an increase in ρ_0). Since the maximum stress intensification in front of a sharp crack occurs over a small region close to the crack tip [26,27], a higher longitudinal tensile stress, σ_{yy} , and thus $K_{applied}$, were required to attain a critical value of the fracture stress, $\sigma_{yy\max} > \sigma_f^*$, over the larger distance. In a strain-controlled situation (such as is apparent for the direct-quenched and tempered 4340 investigated in [19]), the higher austenitizing treatment again resulted in an increased ρ_0 (in this instance, the distance over which it was presumed necessary to attain a critical strain, ϵ_f^*) and thus an increase in K_{IC} . Decreases in Charpy V-notch impact energy, resulting from the high temperature austenitizing treatment, were correlated with decreases in $K_{IC}(\rho)$ for $\rho > \rho_0$, and were attributed to lower intrinsic values of σ_f^* and ϵ_f^* . The lowering of these inherent material parameters was thought to result from impurity segregation at the grain boundaries

(for σ_f^*) [20] and "increased strain concentrations" (which were not explained) at the boundaries of the larger martensite lath colonies (for ϵ_f^*) [19].

In the present work, a simple scheme to rationalize "sharp crack" toughness increases (for a physically reasonable situation involving crack tip blunting) with increased austenitizing temperature, in the as-quenched and 200°C/1 hour temper conditions, is presented. Since regions of dimpled rupture were observed linking the cleaved regions, the fracture mechanism will be presumed to be strain controlled. Further comments on the applicability of the sort of stress-controlled criteria generally found in the literature [24,25,28], to the type of fracture mechanisms observed in this class of steels, will be presented in a later section. We proceed in the following manner: on the basis of finite element studies [23], a reasonably accurate measure of the crack opening displacement, b , is obtained as a function of the intensity of the applied elastic-plastic field, namely

$$b = \frac{\alpha J_{\text{applied}}}{\sigma_{\text{flow}}} , \quad (4.1)$$

where $\alpha \approx .48$ for $\sigma_{ys}/E = 1/200$ and $n = 0.1$, and σ_{flow} is taken to be σ_{ys} . For linear elasticity, and under small-scale yielding conditions in the elastic-plastic regime,

$$J_{\text{applied}} = \frac{K_I^2 (1-\nu^2)}{E} , \quad (4.2)$$

where the factor $(1-\nu^2)$ is included to account for plane strain.

Algebraic manipulation yields the following relation:

$$K_I = \left[\frac{E b \sigma_{ys}}{\alpha (1-\nu^2)} \right]^{1/2} \quad (4.3)$$

The stress and strain distribution along the crack line ahead of a blunted "initially sharp" crack tip, as determined by a finite element study [25], is shown in Figure 4.1. Distance from the crack tip, R , is normalized by the current crack opening displacement, b . At a given value of strain (plastic), a value of the normalized distance from the crack tip, $I(=R/b)$, is readily determined from the figure. At the critical value of strain, ϵ_f^* , the critical intercept, $I^*(=\delta^*/b)$, is obtained. Here the distance R has been identified as the critical distance δ^* over which the critical strain ϵ_f^* must act. It should be noted that ϵ_f^* should be the true fracture strain under nearly plane strain conditions, and not the uniaxial tensile ductility. Substitution of $b = \delta^*/I^*$ into equation (4.3) yields the desired relation:

$$K_{IC} = \left[\frac{E \delta^* \sigma_{ys}}{\alpha (1-\nu^2) I^*} \right]^{1/2} \quad (4.4)$$

Table 4.1 presents the results obtained using equation (4.4). Values of ϵ_f^* were estimated using the results obtained by Clausen [29] on ductility in high strength steels, namely:

$$\frac{\epsilon_f^* \text{ plane strain}}{\epsilon_f^* \text{ uniaxial tension}} \approx 0.24, \quad (4.5)$$

for a steel with a yield strength of 220 ksi. On the basis of the fractographic examination, the lath colony size had been tentatively identified as the fracture controlling parameter δ^* . Mean values of 15 microns and 42 microns for the small and large-grained material, respectively, were used in the calculations of Table 4.1. Although absolute magnitudes of K_{IC} , as predicted by equation (4.4), are too large, significant toughness increase predictions (which are in reasonable accord with those observed experimentally) with higher austenitizing temperature are noted. If, instead, δ^* is taken to vary as the prior austenite grain size, toughness increase predictions are on the order of 130 percent, a value clearly not supported experimentally. We further note that the average lath colony sizes of 6 and 13.3 microns (for the small and large-grained material, respectively) which would yield (using equation (4.4)) the values of K_{IC} found experimentally, are certainly not unreasonable and could well correspond to the lower end of the colony size distribution.

In the work of Ritchie and Horn [19] on direct quenched and tempered 4340, a strain-controlled fracture mechanism was assumed, and an equation for sharp crack toughness,

$$K_{IC} \approx \left[\frac{3}{2} \sigma_{ys} E \epsilon_f^* \right]^{1/2} \rho_o^{1/2}, \quad (4.6)$$

was derived. Values of characteristic distance ρ_o (in our case, δ^*), were determined through slow bend Charpy tests on specimens containing various notch root radii. Although they provided no quantitative metallographic measurements of lath colony (packet) size and/or particle spacing

to correlate with the value of ρ_0 determined by the Charpy test, we note that the values of 17 microns and 48 microns (from [19]) for the material austenitized at 870°C and 1200°C, respectively, are in good accord with the values determined for lath colony size in the present work. This would seem to provide further support for the contention that the fracture process in these temper conditions is colony controlled. Plane strain ductility values ϵ_f^* were calculated using equation (4.6) with values of δ^* (ρ_0) determined in the present work; these are shown in Table 4.1. The values for ϵ_f^* determined in this manner are rather higher than those determined experimentally (using Clausen's result), but the proper trend is observed. Ritchie and Horn [19] observed much the same result in their investigation of direct quenched and tempered 4340, and concluded that a decrease in the intrinsic material parameter, ϵ_f^* , with the increased temperature austenitizing treatments, was the cause of the lower K_A ($\rho; \rho > \rho_0$) and thus Charpy V-notch toughness. They further suggest that the decrease in ϵ_f^* is attributable to "greater strain concentrations developed at the boundaries between the larger martensite packets." This explanation seems rather simplistic in that no specific mechanism for fracture initiation is suggested, and since [19] provides no direct evidence to support it, a more careful evaluation of the reason for the apparent decrease in ϵ_f^* is indicated.

In summary, there is a good deal of evidence, obtained both through fractographic analysis and experimental determinations of ρ_0 (δ^*), to support the contention that the fracture controlling microstructural feature, or microstructural unit, in the as-quenched and 200°C/1 hour

temper conditions, is the lath colonies. A simple scheme, based on the notion that a critical strain must be achieved over a critically large enough distance and utilizing a physically reasonable strain distribution ahead of a blunted crack, provides a good rationalization of K_{IC} increases with higher temperature austenitizing treatments. The notion that a decrement in ϵ_f^* occurs upon higher temperature austenitization provides a good explanation for the decrease in Charpy V-notch toughness *vis-a-vis* the decrease in K_A ($p; p > p_0$). However, the reason for this apparent lowering of ϵ_f^* is not entirely clear, and should be the subject of future investigation.

4.2 Toughness in the 300°C/1 hour and 400°C/1 hour Temper Conditions

Toughness in the 300°C/1 hour and 400°C/1 hour temper conditions, as evidenced by the SEM fractographs and replica electron micrographs presented in section 3, is clearly controlled by the cementite morphology in conjunction with the prior austenite grain size. In the case of the large-grained material, a classic example of 500°F temper embrittlement (tempered martensite embrittlement--TME) is observed. Phenomenologically, toughness decreases (due to tempering) are associated with the concurrent precipitation of cementite and partitioning of various cementite-insoluble elements [30]. These "rejected" elements locally weaken the carbide-ferrite interface, providing for easier void initiation, and ultimate coalescence along the grain boundaries. Experiments on relatively pure heats [31], and on silicon modified 300M steel [32], indicate that both the presence of the impurity elements and the cementite are required for the embrittlement to occur. High temperature grain boundary segregation of impurity elements may act in concert with

the foregoing effect [33]. The replica electron micrographs of the large-grained material show a coarsening and significant agglomeration of the carbides (with a plate-like precipitate along laths and, presumably, also prior austenite grain boundaries) subsequent to tempering at 400°C, where the observed embrittlement, as documented by both the Charpy V-notch and the K_{IC} tests, was greatest. In the 300°C/1 hour temper condition, where the agglomeration and boundary precipitation were not as apparent, the embrittlement (although quite significant with respect to the 200°C/1 hour temper condition) was not as great.

The small-grained material, although exhibiting the embrittlement effect, failed primarily by transgranular quasi-cleavage, and not intergranularly as in the case of the large-grained material. As shown by the replica electron micrographs, tempering at 300°C and 400°C produced a particle sub-structure not unlike that generally observed in upper bainite [34]. Thomas [35], in an investigation of Fe/C/X steels, describes a similar situation. He concludes that the precipitation of inter-lath cementite and alloy carbides (M_3C), as a result of the decomposition of retained austenite, causes the embrittlement. He goes on to state that this embrittlement phenomenon should be distinguished from TME since it also occurs in highly pure alloys, and since it results in a transgranular, and not necessarily intergranular, fracture morphology. In our case, using a commercially-prepared alloy with a significant impurity content, some partitioning concurrent with the carbide precipitation undoubtedly occurred. It is probable, however, that due to the low (870°C) austenitizing temperature and correspondingly

lower diffusion rates, impurity element segregation to prior austenite grain boundaries was not extensive. This, in conjunction with the greater total grain boundary area associated with the smaller grain size, may well have resulted in a less severe grain boundary embrittlement and produced a transgranular cleavage, rather than intergranular, mode of fracture. It is clear however, that the majority of the cleaved regions scale to the prior austenite grain size, and that this should be considered the microstructural unit controlling fracture in the small-grained material (as well as in the more obvious case of the large-grained material).

The situation described herein for the 400°C/1 hour temper condition would appear to be a perfect test case for the sort of stress-controlled criterion applied by Ritchie et al. [20] to direct quenched 870°C and step-quenched (1200°C + 870°C) 4340 structures, where primarily transgranular cleavage and intergranular fractures were observed. In the present work, a temper of 400°C for 1 hour produced fracture morphologies consisting almost wholly of intergranular cracking and transgranular quasi-cleavage, with only limited amounts of fibrosity. As shown by the SEM fractographs, the microstructural unit which would appear to control the fracture process for each austenitizing treatment, is the prior austenite grains. On the basis of the good correspondence between the value of ρ_0 determined by Ritchie and Horn [19] and δ^* as measured in the present work (in each case for strain-controlled fracture processes), we shall presume the value of ρ_0 for the 400°C/1 hour temper condition to vary (i.e., to scale directly proportional to) the

prior austenite grain size in the analysis which follows.

The notion of a stress-controlled criterion is not new and has been examined rather extensively in the literature [24,25,28]. It is based on the slip-line field theory of Hill [36] for elastic-perfectly plastic conditions ahead of a blunt notch, where the maximum longitudinal stress intensification,

$$\frac{\sigma_{yy}^{\max}}{\sigma_{ys}},$$

is located at the elastic-plastic interface. Hill gives the following relation for the maximum longitudinal stress:

$$\sigma_{yy}^{\max} = \sigma_{ys} [1 + \ln(1 + r/\rho)], \quad (4.7)$$

where r is the distance from the notch along the notch line and ρ is the notch root radius. Fracture is taken to occur at some critical value of the plastic zone size $r=r_c$ (with $r_c < \rho [\exp(\frac{\pi-w}{2}) - 1]$ for equation (4.7) to remain valid; see [24]), where r_c is determined from elasticity theory (plasticity is modelled as an equivalent elastic crack [37]) as:

$$r_c \approx .12 \left[\frac{K_A}{\sigma_{ys}} \right]^2. \quad (4.8)$$

At failure then, $\sigma_{yy}^{\max} = \sigma_f^*$, and

$$K_A \approx 2.89 \sigma_{ys} \left[\exp \left(\frac{\sigma_f^*}{\sigma_{ys}} - 1 \right) - 1 \right]^{1/2} \rho^{1/2}. \quad (4.9)$$

As before, in describing a strain-controlled criterion [19], an effective notch root radius, $\rho = \rho_o$, is determined. With appropriate substitution, the following result is obtained:

$$K_{IC} \approx 2.89 \sigma_{ys} \left[\exp \left(\frac{\sigma_f^*}{\sigma_{ys}} - 1 \right) - 1 \right]^{1/2} \rho_o^{1/2}. \quad (4.10)$$

To test the consistency of this criterion, when evaluated using a physically reasonable finite element solution of the stress-distribution ahead of a blunted "initially sharp" crack, we solve for the intrinsic fracture stress σ_f^* as:

$$\sigma_f^* = \sigma_{ys} \left\{ 1 + \ln \left[1 + \left(\frac{K_{IC}}{2.89 \sigma_{ys}} \right)^2 \frac{1}{\rho_o} \right] \right\}. \quad (4.11)$$

At fracture (as defined by the criterion of 2 percent maximum allowable crack growth used in calculating K_{IC}), using experimentally determined values of σ_{ys} and K_{IC} , the maximum stress intensification $\frac{\sigma_f^*}{\sigma_{ys}}$, is 4.3 and 2.5 for the small and large-grained specimens, respectively. As is readily apparent from Figure 4.1, these values lead to physically

unreasonable values of the critical intercept I^* , and hence of δ^* . We further note that, using values of K_{IC} , σ_{ys} , and ρ_0 as reported in [20], values of 2.3 and 1.6 for the maximum stress intensification in the small and large-grained material, respectively, are obtained. These results are also physically unreasonable.

We are thus led to the conclusion that the application of a stress-controlled criterion to the situations described herein, is inappropriate. As noted in both [24] and [28], situations where fracture could be considered to be strictly controlled by the achievement of a sufficiently high longitudinal tensile stress were limited to cleavage at low temperatures ($< -45^\circ\text{C}$). At higher temperatures [such as in the present case], where the initiation of the cleavage micro-cracks was not coincident with catastrophic failure, the authors indicated that the fracture process was strain-controlled and that a stress-controlled criterion was inapplicable.

In summary, fracture toughness in the $300^\circ\text{C}/1$ hour and $400^\circ\text{C}/1$ hour temper conditions is largely controlled by the cementite morphology in conjunction with the prior austenite grain size. In the large-grained material, where the embrittlement (as evidenced by the K_{IC} and Charpy data) was most severe; an intergranular fracture morphology was observed. The small-grained material exhibited a transgranular quasi-cleavage fracture morphology, although in this case as well, the prior austenite grain size was clearly the relevant microstructural unit. Attempts to rationalize the sharp crack fracture toughness, using a stress-controlled fracture criterion, were shown to yield physically unreasonable δ^* values.

In view of past work on low temperature cleavage fracture in mild steel, this result was not surprising, and indicates that a strain-controlled criterion may be more appropriate in the rationalization of ambient temperature "sharp crack" toughness in these steels.

4.3 Toughness and Stable Crack Growth in the 500°/1 hour Temper Condition

Specimens tempered at 500°C for 1 hour were quite tough, and when loaded in the stroke control mode during the J tests, exhibited considerable amounts of stable crack growth. An examination of the SEM fractographs revealed a marked degree of intergranular cracking in the large-grained material (although the overall fracture appearance is one of dimpled rupture) as compared to the virtually 100 percent transgranular ductile rupture in the material austenitized at 870°C. This rationalizes, at least qualitatively, the higher sharp crack toughness (K_{IC} as converted from J_{IC}), and Charpy V-notch impact energy values which were observed for the finer grained specimens.

In this present work, J_{IC} is reported as determined using several criteria (Table 3.4 and Figures 3.21 and 3.22). Values of J_{IC} obtained using the equation

$$\Delta a = \frac{J}{2\sigma_{flow}} \quad (4.12)$$

appear to be the least influenced by the sort of material response variations that one might reasonably expect from similarly prepared and

configured specimens, as well as being relatively specimen size independent [6].

The validity of the J tests was checked using the standard formula

$$b, B > \alpha \frac{J_{IC}}{(2\sigma_{flow})} \quad (4.13)$$

In view of recent findings [38], α was taken to be 100 (for a severe test of validity), as well as the more generally accepted value of 50. As shown in Table 3.4, even for the relatively thin compact tension specimens employed in this work, and for an α of 100, all the J tests were valid.

Paris et al. [11,39] have recently proposed a dimensionless tearing modulus,

$$T = \frac{E}{\sigma_{ys}} \frac{dJ}{da} \quad (4.14)$$

to characterize a growing mode I crack. Under conditions derived in [11], known as J-controlled growth, J as a characterizing parameter continues to be relevant even in the presence of small amounts of crack growth. Higher values of $T_{initial}$ (where $(\frac{dJ}{da})_{initial}$ is the slope of the J resistance curve at the initiation of crack growth) are associated with greater amounts of stable crack growth, and hence, of fracture toughness. Due to the uncertainties of determining the crack initiation point, and hence, of accurately determining $(\frac{dJ}{da})_{initial}$, $\frac{dJ}{da}$ was calculated in the present work for the well-established linear portion of the resistance curve. The good correspondence between T and J_{IC} (and K_{IC}) in the

sense described above, is shown in Table 3.4.

The ridge patterns (or so-called zig-zag crack growth) observed for the 500°C/1 hour temper condition are common to many alloy systems, including OFHC copper, 200 grade maraging steel, 4340 [40], and HY80 steel [38]. A number of theories have been proposed to model this phenomenon; Van Den Avyle [40] for example, suggests that the wavelength of the ridges, λ , varies linearly with COD (and thus with K_{IC}^2). In the present work, an extensive formation of ridge patterns was observed in only one temper condition, so a meaningful analysis of the problem could not be undertaken. We do note, however, that the wavelength of the ridges ($\lambda \approx 110$ microns for the fine-grained compact tension specimens, for example) did not directly correspond to any of the microstructural units identified in this work. This implies that the proper description of ridge formation is not microscopic, but should be based on a continuum analysis [43].

5. CONCLUSIONS

Fracture toughness (K_{IC}) in the as-quenched and 200°C/1 hour temper conditions was found to be rationalized by a criterion which requires that a critical value of strain be achieved over a critically large enough distance. On the basis of fractographic examination, the lath colonies were determined to be the fracture controlling microstructural unit, and hence their size should be considered as the critical distance. The size of the colonies varied from 20 microns to 40 microns for the 870°C and 1175°C austenitizing treatments, respectively. Application of the criterion, using a finite element solution of the strain field ahead of a blunted "initially sharp" crack, yielded toughness increase predictions with increased temperature austenitizing treatment which were in good accord with experiment.

Toughness in the 300°C/1 hour and 400°C/1 hour temper conditions appeared to be largely a function of the cementite morphology and the prior austenite grain size. In the large-grained material, where the mechanical data indicated a severe embrittlement, an intergranular mode of failure was observed. The finer grained material, although embrittled following a temper of 300°C for 1 hour, had recovered subsequent to a 400°C/1 hour temper and exhibited a predominantly transgranular cleavage, and not an intergranular, fracture morphology. Attempts to rationalize toughness in the temper embrittled condition, using a stress-controlled criterion, were shown to yield physically

unreasonable values of the critical distances, and by direct implication, of the fracture toughness as well.

J_{IC} and the tearing modulus, calculated for the 500°C/1 hour temper condition, were in good correspondence as indicators of material toughness. The values obtained for J_{IC} are qualitatively rationalized by the nature of the fracture morphologies, namely, the lower toughness in the large-grained material is attributable to the fact that the material had not fully recovered from the previous severe intergranular embrittlement. In the small-grained material, no obvious intergranular cracking was present, and the fracture morphology was wholly ductile rupture. The observation of ridge patterns in this temper condition is noted; the wavelength of the ridges did not correspond to any of the salient microstructural features, implying the need for a continuum rather than a microscopic description of the fracture process.

6. FUTURE WORK

In order to better elucidate the findings of the present work, further research is being conducted in several areas. Charpy tests are being performed on bars of various notch root radii in an attempt to fully establish the correlation between lath colony size (δ^*) and the critical distance (ρ_0). Further fractographic analysis is being undertaken in order to quantify the salient morphological features. A technique, involving the SEM examination of a polished and etched plane cut at an angle to the fracture surface [14] is being pursued. This should graphically illustrate the interaction of the main crack with the various microstructural features. Along similar lines, analysis of subsidiary cracking, as viewed in longitudinally sectioned Charpy bars and compact tension specimens, is continuing.

The issue raised in section three, regarding the cause of the apparent decrease in tensile ductility and intrinsic plane strain ductility, ϵ_F^* , would seem to be of paramount importance in rationalizing the loss of toughness with increased austenitizing temperature in the blunt notch ($\rho > \rho_0$) case. A further review of the literature is being conducted in an attempt to more precisely identify the mechanics involved in this situation.

Finally, we have hopes of obtaining a heat of titanium killed 4340 steel, wherein the anticipated grain and colony growth with high austenitizing temperatures should be minimal. This will allow us to fur-

ther examine the central thesis of the present work, namely, that the increase in sharp crack fracture toughness with increased austenitizing temperature is largely attributable to increases in the lath colony size (and thus the critical distance). Titanium grain refinement will also allow us to explore the possibly beneficial effects of high austenitizing temperatures with regards to refining the carbide morphologies.

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References

1. Dreyer, G.A., Austin, D.E., and Smith, D.W., Metal Progress, 86, p. 116, (1964).
2. Underwood, E.E., Quantitative Stereology, Chp. 2, p. 23, Addison-Wesley Pub. Co., Reading, Mass., (1970).
3. SEM/TEM Fractography Handbook, p. 9, Metals and Ceramics Information Center, Battelle Columbus Laboratories, Columbus, Ohio, (1975).
4. Annual Book of ASTM Standards, E23-72, p. 167, Am. Soc. for Testing and Materials, Phil., (1975).
5. Annual Book of ASTM Standards, E399-72, p. 1, Am. Soc. for Testing and Materials, Phil., (1972).
6. Landes, J.D., and Begley, J.A., ASTM STP560, p. 170, (1974).
7. Srawley, J.E., NASA Technical Memorandum TMX-71881, presented to Committee E-24 of the Am. Soc. for Testing and Materials, Orlando, Florida, March 22, 1976.
8. Clarke, G., et al., ASTM ST P 590, p. 302, (1976).
9. Hermann, L., Paris, P.C., and Asaro, R.J., to be published.
10. Hermann, L., " K_{open} and Closure Distance Determination in Cyclic Fatigue Tests", Brown University, Unpublished.
11. Hutchinson, J.W., and Paris, P.C., "Stability Analysis of J-Controlled Crack Growth", presented at ASTM Symposium on Elastic-Plastic Fracture, Atlanta, (1977), to be published in ASTM STP.
12. Irwin, G.R., J. Appl. Mech., 24, p. 361, (1957).
13. Cox, T.B., and Low, J.R., Met. Trans., 5, p. 1457, (1974).
14. Almond, E.A., et al., Metallography 3, p. 379, (1970).
15. Lai, G.Y., et al., Met. Trans. 5, p. 1663, (1974).
16. Zackay, V.F., et al., Nature Phys. Sci., 236, p. 108, (1972).

17. Wood, W.E., Eng. Fract. Mech., 7, p. 219, (1975).
18. Wood, W.E., Met. Trans. A, 8A, p. 1195, (1977).
19. Ritchie, R.O., and Horn, R.M., Met. Trans. A, 9A, p. 331, (1978).
20. Ritchie, R.O., et al., Met. Trans. A, 7A, p. 831, (1976).
21. Ritchie, R.O., et al., Met. Trans. A, 8A, p. 1197, (1977).
22. Bernstein, I.M., personal communication.
23. McMeeking, R.M., Brown University Report, Technical Report No. 44, (1976).
24. Malkin, J., and Tetelman, A.S., Engin. Fract. Mech., 3, p. 151, (1971).
25. Wilshaw, T.R., et al., Engin. Fract. Mech., 1, p. 191, (1968).
26. Rice, J.R., and Johnson, M.A., Inelastic Behavior of Solids (eds. Kanninen, M.F., Adler, W.F., Rosenfield, A.R., and Jaffee, R.I.), p. 641, McGraw-Hill, New York, (1970).
27. Rice, J.R. and Rosengren, G.F., J. Mech. Phys. Solids, 16, p. 1, (1968).
28. Ritchie, R.O., Knott, J.F., and Rice, J.R., J. Mech. Phys. Solids, 21, p. 395, (1973).
29. Clausen, D.P., Int. J. of Fract. Mech., 6, p. 71, (1970).
30. Bellick, J.R., and McMahon, C.J., Jr., Met. Trans. 5, p. 2439, (1974).
31. Capus, J.M., and Mayer, G., Metallurgia, 62, p. 133, (1960).
32. Baneji, S.K., et al., Met. Trans. A, 9A, p. 237, (1978).
33. Shulz, B.J., and McMahon, C.J., Jr., ASTM STP 499, p. 104, (1972).
34. Electron Microstructure of Bainite in Steel, Second Progress Report by Subcommittee XI of Committee E-4, Proceedings, Am. Soc. Testing Materials, 52, 543, (1952).

35. Thomas, G., Met. Trans. A., 9A, p. 439, (1978).
36. Hill, R., The Mathematical Theory of Plasticity, Oxford University Press, (1950).
37. Knott, J.F., Fundamentals of Fracture Mechanics, Halsted Press, (1973).
38. Pan, J., Masters Thesis, Brown University, (1978).
39. Paris, P.C., Tada, H., Zahoor, A., and Ernst, H., presented at ASTM Symposium on Elastic-Plastic Fracture, Atlanta, (1977), to be published in ASTM STP.
40. Van Der Avyle, J.A., Ph.D. Dissertation, Massachusetts Institute of Technology, (1975).
41. Tada, H., Paris, P.C., and Irwin, G.R., Del Research Corporation, Pennsylvania, p. 2.20, (1973).
42. Baik, J.M., Masters Thesis, Brown University, (1979).
43. McClintock, F.A., Physics of Strength and Plasticity, (A.S. Argon, ed.), MIT Press, Cambridge, Mass., (1969).
44. P.N. Thielen and M.E. Fine, Met. Trans. A, 6A, p. 2133, (1975).
45. M.-W. Lui and I. LeMay, Met. Trans. A., 6A, p. 583, (1975).
46. H.H. Lee and H.H. Uhlig, Met. Trans., 3, p. 2949, (1972).
47. S. Mostovoy and N.N. Breyer, Transactions of the ASM, 61, p. 221, (1968).

APPENDIX A

The initial crack length, a_0 , was determined in the following manner: first, the initial elastic compliance was calculated as described in Section Two (using equation 2.1). Then, using the result of Tada et al., [41] for a load-line displacement modified compact tension specimen,

$$V(a/w) = C_{\text{ELASTIC}} \frac{E}{(1-\nu^2)} \quad (\text{A.1})$$

in conjunction with their graph of $(1 - a/w)^2 V(a/w)$ versus (a/w) (or the modified form of $V(a/w)$ versus (a/w)), a_0 was obtained.

Crack length changes, da , were determined using the differential compliance scheme outlined in Section Two and certain compliance relations to be derived below [10,42]. Referring to Figure A.1, which depicts the elastic response of an infinite plate containing a crack of length a and $a + \delta a$, respectively, a relation for the energy release rate, G (per unit thickness), is readily obtained as:

$$G B \delta a = \frac{1}{2} P \delta u . \quad (\text{A.2})$$

Here P is the fixed applied load, δa is the increment in crack length, δu is the load point displacement, and B is the plate thickness. Since

$$U = C(a) P , \quad (\text{A.3})$$

where $C(a)$ is the compliance of the plate at a particular value of

crack length, we can write

$$G B \delta a = \frac{1}{2} P^2 \delta C . \quad (A.4)$$

In the limit, as $\delta a \rightarrow 0$, we obtain

$$G = \frac{\frac{1}{2} P^2 \left(\frac{dC}{da} \right)}{B} . \quad (A.5)$$

For plane strain, Mode I loading conditions, and assuming linear elastic material response,

$$G = \frac{1-\nu^2}{E} K_I^2 . \quad (A.6)$$

Equating A.5 and A.6, and performing the required algebraic manipulation, the following relation for crack length change in terms of differential compliance is obtained:

$$da = \frac{E}{2B(1-\nu^2)} \left(\frac{P}{K_I} \right)^2 dC . \quad (A.7)$$

The stress intensity factor K_I , is expressed generally as:

$$K_I = \frac{P}{BW^{1/2}} f(a/w) , \quad (A.8)$$

where $f(a/w)$ is the polynomial expression proposed by Srawley [7]. Substituting A.8 into A.7, we thus obtain the desired relation for crack length change:

$$da = \frac{EB}{2(1-\nu^2)} \left(\frac{W}{f(a/w)} \right)^2 dC . \quad (A.9)$$

TABLE CAPTIONS

- 2.1 Composition of the steel in weight percent, as determined by both a ladle and a billet check.
- 3.1 The inclusion volume fraction for various heat treatments - in percent.
- 3.2 Room temperature, uniaxial tensile properties.
- 3.3 Plane strain fracture toughness (K_{IC}). Pertinent information regarding the validity of each test is included.
- 3.4 J_{IC} values as determined using several criteria. Information regarding the validity of each test is provided.
- 4.1 The predictions of the strain-controlled criterion (equation 4.4) compared with experimentally determined K_{IC} values. Values of ϵ_f^* , as calculated using equation 4.6 [19], are also shown.

FIGURE CAPTIONS

- Figure 2.1 A view of the compact tension specimen and hangar plates. Note the cylindrical hole machined into the specimen and the necked down section of the hangar plates.
- Figure 2.2 A schematic diagram of the electronic system employed in the J-tests.
- Figure 2.3 An example of a typical differential compliance versus load curve, as obtained during a J-test.
- Figure 3.1(a) Optical micrograph of the L-W plane of an as-quenched blank following austenitization at 870°C; etched in Nital.
- (b) Optical micrograph of the L-W plane of a blank austenitized at 870°C and pre-tempered at 350°C for 1 hour. Sodium tridecylbenzene sulfonate was employed as the etchant, to better delineate the prior austenite grain boundaries.
- (c) Optical micrograph of the same blank shown in (b) after an additional etch with Nital. Lath colony orientations within the grains are indicated.
- (d) Optical micrograph of the L-W plane of a blank austenitized at 1175°C and pre-tempered at 200°C for 1 hour. Nital and sodium tridecylbenzene sulfonate were used as etchants.
- (e) Optical micrograph of the L-W plane of an as-quenched blank following austenitization at 1000°C and etching in Nital. A indicates a randomly oriented martensite plate; B indicates a plate on a grain boundary.
- (f) Optical micrograph of the L-W plane of an as-quenched blank following austenitization at 1175°C; etched in Nital.
- Figure 3.2 The effect of increasing austenitizing temperature on the prior austenite grain size and mean lath colony size.
- Figure 3.3(a) Scanning electron fractograph of a MnS inclusion in a compact tension specimen austenitized at 870°C and tempered at 200°C for 1 hour. Note the burn spots from the electron beam.
- (b) Video display of the EDAX results of a spot probe of the particle shown in (a). From left to right, the Sulphur K_{α} , Manganese K_{α} , Iron K_{α} , and Iron K_{β} peaks are prominent.
- (c) Video display of the EDAX results of a spot probe of the matrix shown in (a). Only the Iron K_{α} and K_{β} peaks are significant.

Figure 3.4(a) Optical micrograph of the L-N plane of a hardness determination specimen austenitized at 1175°C and tempered at 200°C for 1 hour. The gray ellipsoidal MnS inclusions are evident against the lightly-etched (Nital and Picral) background of laths.

(b) Optical micrograph of the L-T plane of an as-quenched tensile specimen following austenitization at 870°C. The effect of mill-rolling on the morphology of the MnS inclusion is evident. Note the dark Ca rich particle embedded within the inclusion.

Figure 3.5 Low magnification optical micrograph of the L-T plane of an as-quenched Charpy bar following austenitization at 870°C. The inclusions are shown segregated to the lean zones of the banded structure.

Figure 3.6(a) Transmission electron micrograph of the as-quenched structure following austenitization at 870°C. Precipitates, located exclusively within the laths, are clearly evident.

(b) Transmission electron micrograph of the small-grained structure in the 200°C/1 hour temper condition.

(c)-(e) Transmission electron micrographs of the small-grained structure in the 300°C/1 hour, 400°C/1 hour, and 500°C/1 hour temper conditions, respectively. Lath boundary precipitation, and precipitate coarsening have occurred.

Figure 3.7(a) Transmission electron micrograph representative of the large-grained structure in the as-quenched and 200°C/1 hour temper conditions. Laths containing fine precipitates are interspersed with precipitate-free laths.

(b) High magnification view of the structure of Figure 3.7(a).

(c) Transmission electron micrograph of the large-grained structure following tempering at 300°C for 1 hour. Precipitate coarsening and some lath boundary precipitation have occurred.

(d)-(e) Transmission electron micrograph of the large-grained structure in the 400°C/1 hour and 500°C/1 hour temper condition. Significant precipitate coarsening and lath boundary precipitation have occurred.

Figure 3.8(a) Scanning electron fractograph of a small-grained as-quenched compact tension specimen, showing subsidiary cracking.

(b) Scanning electron fractograph representative of the as-quenched and 200°C/1 hour temper conditions in small-grained Charpy bars and compact tension specimens.

Figure 3.8(c) High magnification view of Figure 3.8(b), showing dimpled rupture and quasi-cleavage fracture morphology.

Figure 3.9(a) Scanning electron fractograph representative of the small-grained compact tension specimens and Charpy bars in the 300°C/1 hour and 400°C/1 hour temper conditions.

(b) High magnification view of Figure 3.9(a), showing increased amounts of cleavage.

(c) Scanning electron fractograph of the ridges (marked by B) observed in the small-grained compact tension specimen tempered at 400°C for 1 hour. The fatigue pre-crack region is denoted by A.

Figure 3.10(a) Scanning electron fractograph of the small-grained compact tension specimen and Charpy bar in the 500°C/1 hour temper condition.

(b) High magnification view of Figure 3.10(a), showing the predominantly dimpled rupture fracture morphology.

Figure 3.11(a) Scanning electron fractograph representative of the large-grained compact tension specimens and Charpy bars in the as-quenched and 200°C/1 hour temper conditions. The fracture morphology consists of mixed transgranular cleavage, dimpled rupture, and some intergranular cracking.

(b) High magnification view of Figure 3.11(a), showing a transgranular cleavage region. These regions, which ranged in size from 30 to 80 microns, were readily apparent in the specimens in the as-quenched and 200°C/1 hour temper conditions.

Figure 3.12 Scanning electron fractograph of the large-grained compact tension specimens and Charpy bars following a temper of 300°C for 1 hour. Increased amounts of cleavage are observed.

Figure 3.13(a) Low and high magnification scanning electron fractographs
(b) of the intergranular cracking characteristic of large-grained specimens in the 400°C/1 hour temper condition.

Figure 3.14(a) Scanning electron fractographs of the large-grained specimens following a temper of 500°C for 1 hour.

(b) High magnification view of Figure 3.14(a), showing the intergranular-dimpled rupture fracture morphology.

- Figure 3.15(a) Optical micrograph of a translatth subsidiary crack in a large-grained, fractured, plated, and longitudinally sectioned compact tension specimen in the 200°C/1 hour temper. Note the change in crack direction with colony orientation. Nital and sodium triclecylbenzene sulfonate were employed as etchants. /
- (b) Optical micrograph of intergranular subsidiary cracking in the same specimen as Figure 3.15(a).
- (c) Fracture surface profile of the specimen of Figure 3.15(a) and (b), showing the smooth intergranular region and rougher translatth (or transcolony) regions.
- (d) Optical micrograph of intergranular subsidiary cracking in a large-grained, longitudinally sectioned Charpy bar in the 200°C/1 hour temper condition; etched in Nital.
- (e) Fracture surface profile of the specimen in Figure 3.15(d), showing the transcolony regions on the surface.
- (f) Optical micrograph of the crack profile, obtained from a pinned and sectioned small-grained compact tension specimen in the 500°C/1 hour temper condition. The well-defined ridges are apparent in the micrograph; unetched.

Figure 3.16 True yield strength (0.2 percent offset), ultimate tensile strength (with load referred to the current area at necking), true fracture strength, and Rockwell hardness as functions of tempering temperature for the 870°C and 1175°C austenitizing treatments.

Figure 3.17 Comparison of uniaxial tensile properties as determined in the present investigation with data obtained by other researchers.

Figure 3.18 True strain to fracture (ϵ_f) and reduction of area (R.A.) as functions of tempering temperature for the 870°C and 1175°C austenitizing treatments. The better ductility of the 870°C treatment is noted.

Figure 3.19 Charpy V-notch impact energy (2 notch orientations) as a function of tempering temperature for low and high temperature austenitizing treatments. Embrittlement is observed at 300°C in the small-grained material and at 300°C and 400°C in the large-grained case.

Figure 3.20 K_{IC} as a function of tempering temperature for the 870°C and 1175°C austenitizing treatments. Data points marked with an asterisk are determined by the following conversion:

$$K_{IC} = \left[\frac{J_{IC} E}{1-\nu^2} \right]^{1/2}$$

- Figure 3.21 J versus da resistance curve for the small-grained specimens tempered at 500°C. Various J_{IC} determination criteria are indicated in the diagram; the intersection of the solid line (corresponding to $\Delta a = J/2\sigma_{flow}$, where $\sigma_{flow} = \sigma_{ys} + \sigma_{ultimate})/2$) with the J versus da curve yielded the value of J_{IC} used in the conversion to K_{IC} .
- Figure 3.22 J versus da resistance curve for the large-grained specimens tempered at 500°C.
- Figure 4.1 Plot of stress (σ_{yy}/σ_{ys}) and plastic strain along the crack line ahead of a blunted "critically sharp" crack. R , the distance from the crack tip, is normalized by the current crack tip opening, b . Critical intercepts are indicated.
- Figure A.1 Schematic load-displacement curve for constant applied loading of an infinite plate containing a crack of length a and $a + \delta a$.

TABLE 2.1

Ladle Check	C	Mn	P	S	Si	Ni	Cr	Mo
	.40	.85	.004	.015	.29	.01	.94	.20
	Cu	V	Al	Sn	Ti	Cb		
	.005	.002	.056	.002	.003	.005		
Billet Check	C	P	S	Mo	As	Sb	Sn	
	.42	.004	.014	.22	.001	.01	.002	

TABLE 3.1

Austenitizing Treatment	Temper					
	As Quenched	200°C	300°C	400°C	500°C	Average
† 870°C	.09	.09	.13	.08	.09	.10
1175°C	.12	.09	.08	.13	.14	.11

TABLE 3.2

	σ_{ys} True		σ_{UTS}		σ_F True		n	ϵ_f Percent	R.A. Percent
	ksi	MPa	ksi	MPa	ksi	MPa			
870°C									
As-quenched	215.1	(1483.4)	331.4	(2285.5)	392.9	(2709.7)	.11	35.1	29.4
200°C/1 hour	226.6	(1562.8)	292.5	(2017.2)	379.3	(2615.9)	.06	53.4	41.4
300°C/1 hour	206.7	(1425.5)	257.7	(1777.2)	318.5	(2196.6)	.05	55.5	42.6
400°C/1 hour	199.1	(1373.1)	227.0	(1565.5)	306.7	(2115.2)	.04	73.9	52.2
500°C/1 hour	167.7	(1156.6)	189.5	(1307.6)	275.7	(1901.4)	.04	83.0	56.5
1175°C									
As-quenched	219.9	(1516.6)	309.9	(2137.2)	359.0	(2475.9)	.10	17.3	15.9
200°C/1 hour	220.5	(1520.7)	291.6	(2011.0)	337.9	(2330.3)	.07	29.9	25.9
300°C/1 hour	207.5	(1431.0)	254.0	(1751.7)	282.6	(1949.0)	.05	29.5	25.6
400°C/1 hour	193.4	(1333.8)	220.6	(1521.4)	243.7	(1680.7)	.03	27.7	24.2
500°C/1 hour	159.9	(1102.8)	184.8	(1274.5)	221.1	(1524.8)	.04	40.5	33.3

TABLE 3.3

	K_{IC}^a		P_{max}/P_Q	$2.5 \left(\frac{K_{IC}}{\sigma_{ys}} \right)^2$	a/w
870°C	ksi- $\sqrt{in}$	MPa- $\sqrt{m}$		mm	
As-quenched	47.7	(52.2)	1.07	3.05	.777
200°C/1 hour	67.5	(73.9)	1.04	5.6	.783
200°C/1 hour	67.5	(73.9)	1.01	5.6	.746
300°C/1 hour	56.8	(62.1)	1.01	4.8	.690
300°C/1 hour	64.0	(70.0)	1.01	6.1	.762
400°C/1 hour	81.9	(89.6)	1.07	10.7	.712
1175°C					
As-quenched	66.4	(72.6)	1.1	5.79	.807
200°C	87.0	(95.2)	1.04	9.88	.762
300°C	73.3	(80.2)	1.04	7.92	.738
400°C	75.1	(82.1)	1.01	9.57	.764

a) corrected for plastic zone size

TABLE 3.4

	J_{IC}^a using $J/2\sigma_{flow}$	J_{IC} using J/σ_{flow}	J_{IC}^c using Δa_c	J_{IC} using 2 Percent Crack Growth	Tearing Modulus ^d (T)	$\alpha J_{IC}/2\sigma_{flow}$ $\alpha=50$ $\alpha=100$	b^e
	lb-in/in ² N-m/m ²	lb-in/in ² N-m/m ²	lb-in/in ² N-m/m ²	lb-in/in ² N-m/m ²		mm mm	mm
1 ^b -52	315 (5.52)	350 (6.13)	445 (7.79)	445 (7.79)	6.02	2.24 4.47	14.9
1-53	385 (6.74)	420 (7.36)	540 (9.46)	540 (9.46)	8.02	2.74 5.49	13.3
2 ^b -4	138 (2.42)	139 (2.43)	142 (2.49)	147 (2.57)	.6	.91 1.80	15.5
2-50	248 (4.34)	265 (4.64)	325 (5.69)	345 (6.04)	5.4	1.85 3.71	12.2
2-51	300 (5.25)	320 (5.60)	425 (7.44)	445 (7.79)	8.3	2.24 4.47	14.1
2-52	285 (4.99)	305 (5.34)	370 (6.48)	385 (6.74)	5.4	2.13 4.27	13.6

a) Value of J_{IC} used in conversion to K_{IC} ($= \{ \frac{J_{IC} E}{1-\nu^2} \}^{1/2}$)

b) 1 and 2 denote specimens austenitized at 870°C and 1175°C, respectively.

c) $\Delta a_c = \frac{.025 E J_{IC}}{\sigma_{ys}^2 (1-\nu^2)}$, $\nu = .3$

d) $T = \frac{\Delta J}{\Delta a} \frac{E}{\sigma_{ys}^2}$, $E = 2 \times 10^5$ MPa

e) b is the remaining ligament of the fatigue pre-crack. The thickness of all the specimens is 11.7 mm.

TABLE 4.1

	ϵ_f^* uniaxial	ϵ_f^* (a) plane strain	K_{IC} theoretical	K_{IC} measured	ϵ_f^* calculated using δ equation (4.6) and δ from present work
870°C	Percent	Percent	ksi- $\sqrt{\text{in.}}$	ksi- $\sqrt{\text{in.}}$	Percent
As quenched	35.1	8.4	85.6	47.7	41
200°C	53.4	12.8	96.6	67.5	79
1175°C					
As quenched	17.3	4.2	128 $\left(\begin{smallmatrix} +49^{(b)} \\ \text{percent} \end{smallmatrix} \right)$	66.4 $\left(\begin{smallmatrix} +39 \\ \text{percent} \end{smallmatrix} \right)$	28
200°C	29.9	7.1	138 $\left(\begin{smallmatrix} +43 \\ \text{percent} \end{smallmatrix} \right)$	87.0 $\left(\begin{smallmatrix} +28 \\ \text{percent} \end{smallmatrix} \right)$	48

a) As determined using Clausing's [30] result of ϵ_f^* plane strain $\approx 0.24 \epsilon_f^*$ uniaxial tensile.

b) Percent increase in K_{IC} from the 870°C austenitizing treatment to the 1175°C austenitizing treatment.

Figure 2.1

Figure 2.3

Figure 2.2

20 μm

(a)

(b)

20 μm

13.3 μm

(c)

(d)

100 μm

Figure 3.1

Figure 3.2

20 μ m

(e)

20 μ m

(f)

Figure 3.1

10 μ m

(a)

(b)

Figure 3.3

Figure 3.3c

Figure 3.4a

10 μ m

6.7 μ m

Figure 3.4b

Figure 3.5

33.3 μ m

(a)

(b)

(c)

(d)

Figure 3.6

Figure 3.6e

1.4 μm
└───┘

1.4 μm
└───┘

Figure 3.7a

0.5 μm

(b)

1.4 μm

(c)

1.4 μm

(d)

Figure 3.7

Figure 3.7e

1.4 μm

192 μm

(a)

77 μm

(b)

Figure 3.8

14.5 μm Figure 3.8c

Figure 3.9a 81.3 μm

14.3 μm

(b)

93.5 μm

(c)

Figure 3.9

77 μm

(a)

15 μm

(b)

Figure 3.10

76 μm

(a)

9 μm

(b)

Figure 3.11

151.5 μm

Figure 3.12

Figure 3.13a

151.5 μm

31.3 μm

Figure 3.13b

Figure 3.14a

151.5 μm

41.5 μm
└───┘

Figure 3.14b

20 μm
└───┘

Figure 3.15a

20 μm

(b)

20 μm

(c)

Figure 3.15

(d)

(e)

33.3 μm

Figure 3.15

Figure 3.15 f

Figure 3.16

TRUE YIELD STRENGTH, ULTIMATE TENSILE STRENGTH,
TRUE FRACTURE STRENGTH, ksi

Figure 3.17

Figure 3.18

Figure 3.19

Figure 3.20

Figure 3.21

Figure 3.22

Figure 4.1

Figure A.1