The effect of microstructure on the cleavage strength of quenched and tempered steels

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Summary
A series of carbon-manganese steels, containing 0.18 to 0.81% carbon and 0 to 1.5% manganese were austenitized, quenched, and tempered. These compositions and heat treatments allowed a variation of maximum carbide size from 0.2 to 5 microns and a variation of free interparticle spacing from 0.13 to 3.0 microns.

Standard Charpy V-notch specimens were fractured in slow three-point bending over a range of temperatures from −196°C to −75°C where fracture occurred by cleavage. Companion tensile tests were conducted over the same temperature range to determine the effect of microstructure on the tensile yield strength, \( \sigma_y^* \).

The critical tensile stress required to cause cleavage fracture below the notch root, \( \sigma_y^* \), was computed by elastic-plastic analyses, wherein \( \sigma_y^* = 2.95 \sigma_y \) at the point where \( P_F = 0.08 P_{GY} \), \( P_F \) is the notched fracture load and \( P_{GY} \) the (extrapolated) value of the general yield load.

The values of \( \sigma_y^* \) varied from 142 to 335 ksi as the microstructure was refined. These values were quantitatively compared with theoretical values for single phase and dispersed systems. The data indicate that in fine microstructures \( \sigma_y^* \) is determined primarily by the free interparticle spacing. In coarse microstructures, particles may act as Griffith-type flaws if the free ferrite path length is sufficient to permit pile-ups to crack the particles; \( \sigma_y^* \) is then determined by the maximum size of the carbide particles. The results also indicate that the presence of a few large particles may not cause a decrease in cleavage strength if the microstructure is refined by a homogeneous distribution of smaller particles.

Introduction
The effect of low volume fractions (<15%) of hard particles on the low temperature cleavage processes in dispersed systems such as spheroidized steel has received little systematic study. From a series of isolated experiments on a variety of materials tested at temperatures where the matrix is capable of failing by cleavage, it appears that the particles can assist or inhibit the cleavage process in the following ways:

1. At small plastic strains the particles may crack or separate from the matrix, creating Griffith-type flaws which are more effective stress concentrators for promoting matrix cleavage than are blocked slip bands [1]. The cleavage strength, \( \sigma_y^* \), is then reduced by the introduction of the large particles

2. A network of thin particles at matrix grain boundaries may effectively harden the boundaries and inhibit plastic relaxation around blocked slip bands and incipient microcracks [2]. The cleavage strength is then reduced by the introduction of the particles [3].
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3. The particles may effectively block microcracks which have formed in the matrix, limiting their size to that of the particle spacing [4]. The cleavage strength is then increased by the introduction of the particles.

4. The particles may block slip processes in the matrix, reducing the length of dislocation pile-ups which initiate fracture and thereby increase the cleavage strength of the system [5].

5. The particles may have no direct effect on the fracture process, but may increase the cleavage strength by causing the matrix grain size to be refined during heat treatment.

6. Finally, on a macroscopic scale, the particles may reduce the triaxial stress state around a notch if the particle-matrix interfaces are weak in shear, by promoting a more homogeneous deformation pattern in the vicinity of the notch [6]. This reduction in triaxiality leads to an increase in macro toughness parameters such as $K_{IC}$ and a reduction in nil-ductility temperature.

It seems certain that the effect of particles on cleavage fracture is not the same in all dispersed systems. In fact, in any one system two or more of these processes may be occurring simultaneously. This is particularly the case in a commercial material that contains more than one type of particle, or in which there exists a wide distribution of particle sizes.

The present investigation was performed to quantitatively determine the effect of particle size, spacing and volume fraction on the microscopic cleavage fracture strength of a system containing low volume fractions of hard, spherical particles which are strongly bonded to the matrix but which are not coherent with it. It was determined that for the spheroidized carbon-manganese steel investigated here, the predominant effects of particles on cleavage fracture are those described above as effects (1) and (4).

**Experimental procedure**

A series of carbon and carbon-manganese steels were chosen as the subject for this investigation. The compositions are given in Table I.

The steels were austenitized to yield an austenite grain size of ASTM No. 3 (0.1 mm), quenched in oil and tempered in static vacuum at temperatures from 450 to 715°C for 24 hours. Typical structures of one steel tempered at three temperatures are shown in Fig. 1. The microstructural parameters were measured from electron microscopy replicas of polished and electro-etched surfaces and the structures were characterized using the following formulas [7].

\[
\lambda_p = \frac{1-f}{N_{pa}}; \quad d = \sqrt{\frac{3}{2}} \frac{\lambda_p f}{1-f} \quad \text{and} \quad D_s = \frac{3}{2} \sqrt{\frac{\pi}{I}}
\]

**Experimental results**

The microstructural parameters determined for each structure are given in Table II. The values for the double-quenched-and-tempered structure are extrapolated from the higher tempering temperatures for steel 81M as the structure was too fine to accurately measure any values but $d_{max}$, the maximum particle size. For all other structures, the calculated values of $d$, the mean particle size, were in good agreement with values estimated
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from the micrographs, leading to confidence in the numerical description of the microstructures.

The ambient temperature tensile yield strength of each microstructure is given in Table 2. Even at liquid nitrogen temperature, plane stress
tensile specimens of these steels fracture by ductile rupture, which
occurs at tensile stress levels well below that required to cause cleavage.
In order to study the effect of microstructure on cleavage strength, it was
therefore necessary to use bars of Charpy V-notch geometry in which the
tensile stress level can be raised above the tensile yield strength, \( \sigma_Y \),
by the triaxial constraint below the notch. The detailed procedure of
determining the fracture strength is described in reference [2], and is
outlined in Fig. 4. Briefly, it has been shown [9, 10] that when the
applied load, \( P \), is 0.8 of the load required for general yielding, \( P_{0.8Y} \),
the maximum tensile stress level, \( \sigma_{SV} \), in the plastic zone ahead of
the V-notch in a perfectly plastic solid is about 1.95 \( \sigma_Y \). This occurs at a
point located in the plane of the notch, roughly two notch root radii ahead
of the notch tip, and is the site of fracture initiation. The general yield
load curve (Fig. 4) is extrapolated a bit below the temperature \( T_{DAX} \),
(where fracture occurs upon general yield) to the temperature \( T^* \) where
\( P_{0.8Y} = 0.8P_{0.8Y} \). From companion tensile tests, the value of \( \sigma_Y \) is
determined at \( T^* \) so that the maximum stress level in the plastic zone
at fracture (i.e., the fracture strength) is given by

\[ \sigma_f^* = 1.95 \sigma_Y \]  

Although this approach is strictly valid for non-strain hardening solids,
it probably gives a good approximation of the effect of relative changes
in microstructure on fracture strength in a given alloy system having
similar strain hardening characteristics. There is abundant evidence
[11, 12] to show that the values of \( \sigma_f^* \) determined by this type of an
approach are roughly independent of temperature provided there is no
change in the mode of fracture initiation, such as from twinning to slip.
Since all of our fractures were slip-nucleated, the fact that \( \sigma_f^* \) for the
various microstructures was determined at different temperatures (see
Table 2) is no hindrance in comparing the relative values.

The major theories of yielding and fracture in dispersed systems pre-
pdict that the yield strength should vary as the reciprocal of the particle
spacing (Fig. 5) or as the reciprocal square root of some ‘effective grain
size’ (Fig. 6). The fracture stress should vary, if theory is obeyed, as the
reciprocal square root of this same effective grain (Fig. 6) or as the
reciprocal square root of the maximum particle size (Fig. 7). The yield
and fracture strength values from Table 2 were therefore plotted against the
appropriate microstructural parameters. It is seen that the yield
strength varies linearly with the reciprocal square root of the free inter-

Discussion

A number of theories have been proposed to explain the yield behavior
of dispersion-strengthened systems. In simple systems a basic mechanism
such as Orowan bowing [13] often controls yielding; in more complex
cases where grain boundaries and substructure are present this is not
the case. Spheroidized steels are typical of this latter category. In
agreement with Liu [14] and Mima [15], the results of this study show
that the yield strength is a linear function of the reciprocal square root
of some measure of dispersion spacing. The least scatter was found when
plotting the yield strength, \( \sigma_f^* \), against the reciprocal square root of the
free interparticle spacing, \( (D_x - D)^{-1/2} \). This is seen in Fig. 6, as opposed
to Orowan’s prediction in Fig. 5.

The yield results are in accord with the conclusions of Turkalo [16]
and Embury, Keh, and Fisher [17]. Turkalo used transmission electron
microscopy to show that dislocation cell walls, or sub-grain boundaries
exist between the particles in spheroidite. The particles thus act as
‘corners’ to pin a sub-grain structure of the same size as the interparticle
spacing. Embury et al. showed that continued cold working refined the
sub-grain structure, or cell size, in rolled and drawn pearlite. They
found that the flow stress varied as the reciprocal square root of this cell
size. The linear relation obtained by Embury et al. is plotted in Fig. 6
and it is seen that their values for flow stress coincide almost exactly
with the values for yield stress in the spheroidite. This indicates that the
chosen measure of the ‘effective grain size’ in spheroidite is correct
and the particles are influencing yielding only as they serve to pin the
dislocation substructure during the tempering treatment.

The initiation and unstable propagation of microcracks through the
large carbide particles should obey the same general laws as those which
describe microcrack initiation and initial propagation in a single-phase
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material. According to the Cottrell-Petch model, for example, the tensile stress, $\sigma_t$, required to cause cracks to spread through the carbides would be

$$\sigma_t = \frac{2 G_c}{nb} = \frac{4 G_y e}{D_k (D_k - d)^{1/2}} \quad (3)$$

where $G_c$ is the shear modulus of the carbides, $G_y$ is the work required for propagation through the carbide, $n$ is the number of dislocations that have piled up at the carbide-matrix interface, $b$ is the Burgers’ vector of the dislocations, and $k$ is the Petch parameter that describes the variation of yield stress with $(D_k - d)/2$ in Fig. 6. In plane tensile specimens, large plastic strains would be required to develop cracks that can propagate the present investigation, however, the local tensile stresses in the plastic zone reach 1.95 $\sigma_y$, which probably accounts for the development of carbide cracks at small plastic strains. An example of such a crack is shown in Fig. 3.

If carbide cracks are, in turn, able to propagate into the ferrite and cause the observed matrix cleavage, then the cleavage strength, $\sigma_c^*$, will be given either by $\sigma_t$ or by the stress required for carbide crack propagation into the matrix, $\sigma_p$, whichever is the greater. The stress $\sigma_p$ is given by

$$\sigma_p = \left[ \frac{4 E y_f}{\pi(1 - \nu^2)} \right]^{1/2} d^{-1/2} \quad (4)$$

where $E$ is the elastic modulus, $y$ is the effective ferrite surface energy, $\nu$ is Poisson’s ratio and $d$ is the carbide particle size. The presence of non-propagating carbide cracks (Fig. 3) suggests that $\sigma_p > \sigma_t$, for structures containing large carbides. Furthermore, the fact that $\sigma_c^*$ increases with decreasing $d_{max}$ even when $(D_k - d)$ is maintained constant (compare steels 81M and 44, 73 and 44MS in Table 2) suggests that equation (4) rather than (3) is controlling and that $\sigma_c^* = \sigma_p$ should vary linearly with $d_{max}$. As seen in Fig. 7, this is roughly true for those structures containing carbides larger than one micron in diameter. The value of $y_f$, the work done in propagating a microcrack through the ferrite, is $1.25 \times 10^4$ ergs/cm$^2$, which is similar to that reported by Low [18] for crack propagation in ferrite.

When both large particles and a fine microstructure are present, as in the double-quenched-and-tempered structure, then $\sigma_c$ can be greater than $\sigma_p$, and $\sigma_c^*$ is not given by the value of $\sigma_p$. As shown in Fig. 7, $\sigma_c^*$ for the DQT structure is 50% greater than predicted by a fracture criterion based simply on the propagation of cracks out of the large carbide par-

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2. The cleavage fracture strength obeys a Griffith criterion for failure if the following conditions hold.
   (a) Carbide particles approximately one micron in diameter or larger are present.
   (b) The matrix structure is sufficiently coarse to allow slip bands to crack the particles.

3. If 2 above is not obeyed, then the fracture strength varies as the inverse square root of the planar interparticle spacing.

4. Neither the yield nor the fracture strength varies directly with the volume fraction of carbide, but this parameter is important insofar as it affects the interparticle spacing.

Acknowledgments
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References


Table 1

<table>
<thead>
<tr>
<th>Steel*</th>
<th>Carbon (w%)</th>
<th>Mn (w%)</th>
<th>S (w%)</th>
<th>All others</th>
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* Note that the steel designation indicates the amount of carbon and whether the steel contains manganese or sulfur in significant amounts.

Table 2

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<th>Steel</th>
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<th>( \lambda_p )</th>
<th>( \lambda_p ) particle size microns</th>
<th>Free interparticle spacing microns</th>
<th>( \Delta_{max} ) Maximum observed particle size, microns</th>
<th>( \sigma_\gamma ) Lower yield stress (323°C), ksi</th>
<th>( \epsilon_f ) Cleavage fracture stress ksi</th>
<th>( T(D/N) ) Notch ductility temperature °C</th>
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Fig. 1. Typical microstructures of quenched and tempered steels. Steel 44M tempered at (a) 700 °C, (b) 590 °C, and (c) 500 °C, respectively. Electro-etch in perchloric-acetic acid solution.

Fig. 2. Steel 81M double-quenched-and tempered, last temper at 450°C.

Fig. 3. Microcrack in carbide occurring at low strain (<1%) in region of highest stress below a notch. Steel 60, 712°C temper.

Fig. 4. Typical curves of slow-bend Charpy general yield load and fracture vs. test temperature (Steel 44M, 700°C temper).

Fig. 5. Room temperature stress plotted against the reciprocal of the mean free particle spacing.
Fig. 6. Room temperature yield stress and cleavage fracture stress vs. the reciprocal square root of the free interparticle spacing.

Fig. 7. Cleavage fracture stress vs. the reciprocal square root of the diameter of the largest carbide particles in a structure.

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