ON THE FRACTURE TOUGHNESS OF WC-Co CEMENTED CARBIDES

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INTRODUCTION

Cemented carbides develop their best mechanical properties in compression. When subjected to tensile stresses, the fracture stress of at least some alloys seems to be determined by the size of the internal and surface defects. [1,2]. The material parameter that determines the resistance to crack propagation from these defects, the fracture toughness, has consequently received considerable attention, and plane strain fracture toughness data have been reported by different investigators, [3,4,5]. It has been found that $K_{IC}$ increases with increasing binder content and carbide crystal size. In the present paper some additional $K_{IC}$ measurements are reported and a model is proposed, which gives a qualitative explanation for the variation of $K_{IC}$ with microstructure.

EXPERIMENTAL

Five alloys were investigated. From scanning electron micrographs of polished sections the volume fractions of carbide, $f_d$, and binder, $f_b$, were determined by point-counting. The numbers of carbide grains intersected, $N_d$, and binder layers traversed, $N_b$, by a line of length $l_{av}$ was determined. From this data the carbide crystal size, $d_d = f_d l_{av}/N_d$, and the mean binder layer thickness, $d_b = f_b l_{av}/N_b$, were calculated. The fracture toughness was determined according to the method previously described by Ingelstrom and Nordberg [5]. A sharp pre-crack was introduced into each compact-tension specimen by a wedge-impact method. The results are summarized in Table 1.

DISCUSSION

Consider the situation at the tip of a crack stressed such that it is on the verge of propagating. The crack may terminate at a carbide grain, or at a binder layer as shown in Figure 1, or between these two points. We will assume that the crack terminates at a binder layer as shown to the right in Figure 1. This assumption is supported by the following argument: The fracture toughness of WC-Co cemented carbides extrapolates to approximately 5 MPa m$^{1/2}$ in the limit $f_b = 0$, [5], which, incidentally, is slightly higher than the values reported for substances as Al$_2$O$_3$ and SiC [6]. Comparing this with a typical figure of 10 MPa m$^{1/2}$ for cemented carbides, we note that the strain energy released on crack advance in the composite by far exceeds the energy consumed when a carbide grain is cracked. It is thus energetically favourable to crack all the carbide grains lying immediately ahead of the pre-crack front at a stress level lower than that which causes overall crack propagation. This, together with the observation that

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the fracture toughness is largely determined by the mean binder layer
thickness, [3] and [4], leads us to examine the situation in the binder
layer ahead of the crack tip.

It has been shown by Sarin and Johansson, [7], that plastic deformation
of the binder phase in WC-Co cemented carbides occurs via a f.c.c.-c.p.h.
Shockey partial dislocation on every second close-packed plane. They
compatibility demands become increasingly hard to meet as deformation pro-
ceeds, and when a certain strain level is reached the binder cracks, Figure
2. We will apply this idea in formulating a criterion for crack propaga-
distance, [8] and [9], which in the present case is the mean binder layer
propagates when the strain in the binder layer, exceeding a distance of d_b
the strain ahead of the crack tip, we use an expression due to Rice and
with a hardening exponent n. Neglecting the angular dependence, the strain
at a distance x from the crack tip is:

\[
e(x) \propto \frac{1}{x^{1+n}}
\]

Normalizing this expression with respect to the strain at the elastic-
modulus, \( e_0 = \sigma_0 / E_c \) (where \( \sigma_0 \) is the flow stress and \( E_c \) the
plastic interface), \( e_0 = \sigma_0 / E_c \), we find that

\[
e(x) = \frac{1}{e_0} \left( \frac{K_{IC}^2}{6n \sigma_0^2} x \right)^{1-n}
\]

The strain in the composite is mainly confined to the binder, so we assume
that the strain in the binder is given by

\[
e_B = \frac{e}{e_B}
\]

Applying the above criterion we set \( e_B = e_1 \) and \( x = d_B = \frac{K_{IC}^2}{6n \sigma_0^2} \) \( e_0 \sigma_0^2 \), and equations (2) and
(3) give that

\[
K_{IC}^2 = 6n \sigma_0^2 e_0 (e_B e_c / \sigma_c)^{n+1}
\]

In view of the data concerning the compressive deformation of WC-Co cemented
 carbides reported by Doi et al., [11] we set \( n = 0.25 \). Setting \( \sigma_0 \) equal
to the 0.2% compressive flow stress, we are in a position to compare experi-
mental data to those calculated with the aid of equation (4), Figure 3. In the work by Ingelström and Nordberg [5], \( d_B \) is not re-
quantiies were calculated using the expressions in reference [12]. The
missing systematic deviation of the KIC data due to Lueth can be explained by the
slope of the straight line in Figure 3 corresponds to \( e_B = e_1 = 3.7 \%, a figure
which seems quite reasonable in view of the estimate of the strains in-
olved in the f.c.c.-c.p.h. transformation made in Ref. [7], 68.

Part I - Physical Metallurgy

Equipped with the above analytical expression for \( K_{IC} \) and the relation for
the 0.2% flow stress and \( d_B \) in Ref. [12] we can plot the relation between
fracture toughness and flow stress for alloys with different carbide grain
sizes, Figure 4. This diagram gives a rationalization for the fact that
commercial high-strength alloys are fine-grained, while the low-strength
alloys are coarse-grained.

The strength of cemented carbides is usually determined in a transverse
rupture test. When the transverse rupture strength, \( \sigma_T \), is plotted as a
function of \( d_B \), as in the work by Gurland [13], it is found that for each
alloy composition, the curve consists of an ascending part for small values of
\( d_B \) and a descending part for large values and that \( \sigma_T \) has a maximum in
the neighbourhood of \( d_B = 1 \, \mu m \). We will now try to reproduce this plot.
For brittle high-strength alloys where no discernible plastic deformation
precedes fracture, we postulate that the mean size of the inherent defects, a,
determines the fracture strength:

\[
\sigma_{T}^{(1)} = \frac{K_{IC}}{\sqrt{2a}}
\]

In the more ductile alloys these defects may be too small to cause fracture
before plasticity (provided of course that a does not change with micro-
structure). In this range we set

\[
\sigma_{T}^{(2)} = \frac{d_B}{\sqrt{2a}}
\]

with \( \sigma_0 \) equal to the 0.2% compressive flow stress. In Figure 5
\( \sigma_{T} = \min (\sigma_T^{(1)}, \sigma_T^{(2)}) \)
is plotted as a function of \( d_B \) for the case of a = 10 \, \mu m . It may be noted
that the main features of Gurland's plot are reproduced.

CONCLUSIONS

A model has been developed for the fracture toughness of WC-Co cemented
 carbides, which relates fracture toughness to microstructural data.

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### Table I

Results of quantitative metallographic measurements and fracture toughness determination, the error in $K_I$ is the variance in a set of twelve measurements. $E_C$ and $\sigma_C$ are the elastic modulus and 0.2% compressive flow stress quoted by the manufacturer.

<table>
<thead>
<tr>
<th>Alloy Number</th>
<th>$r_\beta$</th>
<th>$d_{\alpha'}$</th>
<th>$d_B$</th>
<th>$K_{IC}$ MPa</th>
<th>$E_C$ MPa</th>
<th>$\sigma_C$ MPa</th>
</tr>
</thead>
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<tr>
<td>1</td>
<td>0.10</td>
<td>3.0</td>
<td>0.70</td>
<td>10.8±0.3</td>
<td>6.5x10^4</td>
<td>3500</td>
</tr>
<tr>
<td>2</td>
<td>0.10</td>
<td>1.1</td>
<td>0.25</td>
<td>9.2±0.4</td>
<td>6.5x10^4</td>
<td>4300</td>
</tr>
<tr>
<td>3</td>
<td>0.15</td>
<td>2.9</td>
<td>0.95</td>
<td>17.3±0.9</td>
<td>5.9x10^5</td>
<td>2700</td>
</tr>
<tr>
<td>4</td>
<td>0.25</td>
<td>3.0</td>
<td>1.90</td>
<td>23.1±2.4</td>
<td>5.4x10^5</td>
<td>2100</td>
</tr>
<tr>
<td>5</td>
<td>0.23</td>
<td>0.95</td>
<td>0.35</td>
<td>11.9±2.4</td>
<td>5.4x10^5</td>
<td>2700</td>
</tr>
</tbody>
</table>

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**Figure 1** Two possible crack tip configurations.

**Figure 2** Transmission electron micrograph of a WC-15%Co alloy, deformed by a Vickers pyramid loaded with 50N. The micrograph was taken from an area located approximately 0.2 mm from the indentation edge. The features in the binder intersecting at the crack show a contrast which is similar to the c.p.h. lamellae shown by Sarin and Johansson [7].

218
Figure 3  $K_{IC}$ as a function of microstructural parameters.

Figure 4  Calculated relation between $K_{IC}$ and 0.2% flow stress for alloys with different carbide grain sizes.

Figure 5  Estimated fracture stress as a function of $d_\beta$. 