Fracture 1977, Volume 2, ICF4, Waterloo, Canada, June 19 - 24, 1977

ON THE FRACTURE TOUGHNESS OF WC-Co CEMENTED CARBIDES

L. Lindau*

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_α, and binder, f_β, were determined by point-counting. The numbers of carbide grains intersected, N_α, and binder layers traversed, N_β, by a line of length L_0 was determined. From this data the carbide crystal size, $d_\alpha = f_\alpha L_0/N_\alpha$, and the mean binder layer thickness, $d_\beta - f_\beta L_0/N_\beta$, 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 I.

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 $MPam^{1/2}$ in the limit $f_\beta = 0$, [3], which, incidentally, is slightly higher than the values reported for substances as Al_2O_3 and SiC [6]. Comparing this with a typical figure of 10 $MPam^{1/2}$ 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

*Department of Physics, Chalmers University of Technology, Göteborg, Sweden.

 MS310

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 Johannesson, [7], that plastic deformation of the binder phase in WC-Co cemented carbides occurs via a f.c.c.→c.p.h. transformation; a process which can be envisaged as the passage of a Shockley partial dislocation on every second close-packed plane. They also argued that, since there is only one slip-plane in the c.p.h. lamellae, compatibility demands become increasingly hard to meet as deformation proceeds, and when a certain strain level is reached the binder cracks, Figure 2. We will apply this idea in formulating a criterion for crack propagation. Such a criterion should also specify a microstructurally significant distance, [8] and [9], which in the present case is the mean binder layer thickness, d_β. The criterion for crack propagation thus reads: the crack propagates when the strain in the binder layer, extending a distance d_β from the crack tip, has reached a certain strain level, ε_1. To calculate the strain ahead of the crack tip, we use an expression due to Rice and Rosengren, [10], which applies to a power law strain-hardening material with a hardening exponent n. Neglecting the angular dependence, the strain at a distance x from the crack tip is:

$$\varepsilon(x) \propto x^{-\frac{1}{1+n}} \tag{1}$$

Normalizing this expression with respect to the strain at the elastic-plastic interface, $\varepsilon_o = \sigma_c/E_c$ (where σ_c is the flow stress and E_c the modulus), and the size of the plastic zone, $x_o = K_{IC}^2/6\pi\,\sigma_c^2$, we find that

$$\varepsilon(x) = \varepsilon_o\,(K_{IC}^2/6\pi\sigma_o^2\,x)^{\frac{1}{1+n}} \tag{2}$$

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

$$\varepsilon_\beta = \varepsilon/f_\beta \tag{3}$$

Applying the above criterion we set $\varepsilon_\beta = \varepsilon_1$ and $x = d_\beta$ and equations (2) and (3) give that

$$K_{IC}^2 = 6\pi d_\beta\,\sigma_c^2\,(\varepsilon_1 f_\beta E_c/\sigma_c)^{n+1} \tag{4}$$

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 σ_c equal to the 0.2% compressive flow stress we are in a position to compare experimentally determined K_{IC} data to those calculated with the aid of equation (4), Figure 3. In the work by Ingelström and Nordberg [5], d_β is not reported, as is the case for σ_c in the work by Leuth [4]. These missing quantities were calculated using the expressions in reference [12]. The systematic deviation of the K_{IC} data due to Lueth can be explained by the less sophisticated method employed for introducing the pre-crack [5]. The slope of the broken line in Figure 3 corresponds to $\varepsilon_1 = 3.7\%$, a figure which seems quite reasonable in view of the estimate of the strains involved in the f.c.c.→c.p.h. transformation made in Ref. [7], 6%.

Equipped with the above analytical expression for K_{IC} and the relation for the 0.2% flow stress and d_β 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, σ_f, is plotted as a function of d_β, 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_β and a descending part for large values and that σ_f has a maximum in the neighbourhood of $d_\beta = 1$ µm. 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:

$$\sigma_f^{(1)} = K_{IC}/\sqrt{\pi a} \tag{5}$$

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 microstructure). In this range we set

$$\sigma_f^{(2)} = \sigma_c \tag{6}$$

with σ_c equal to the 0.2% compressive flow stress. In Figure 5

$$\sigma_f = \min\,(\sigma_f^{(1)}\ ,\ \sigma_f^{(2)})$$

is plotted as a function of d_β for the case of a = 10 µm. 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.

ACKNOWLEDGEMENT

This work was supported by the Swedish Board for Technical Development. Alloy preparation and fracture toughness measurements were carried out at the Coromant Research Centre, Sandvik AB.

REFERENCES

1. SUZUKI, H. and HAYASHI, K., Planseeber. f. Pulvermet., 23, 1975, 24.
2. EXNER, H., WALTER, A. and PABST, R., Mat. Sc. Engng., 16, 1974, 231.
3. OSTERSTOCK, F., Diplôme d'Etudes Approfondies, Université de Caen, 1974.
4. LUETH, R. C., Fracture Mechanics of Ceramics, II, R.C. Bradt, D. P. H. Hasselman and F. F. Lange (eds.), New York, 1974, 791.
5. INGELSTRÖM N. and NORDBERG, H., Engng. Fracture Mech., 6 1974, 597.
6. Data quoted in KIRCHNER, H. P., GRUVER, R. M. and SOTTER, W. A., Phil. Mag. 33, 1976, 775.
7. SARIN, V. K. and JOHANNESSON, T., Met. Sci., 9, 1975, 472.

8. RICE, J. R. and JOHNSON, M. A., Inelastic Behaviour of Solids, M. F. Kanninen, W. Adler, W. Rosenfield and R. Jaffe (eds.), New York, 1970, 641.
9. RITCHIE, R. O., KNOTT, J. F. and RICE, J. R., J. Mech. Phys. Solids, 21, 1973, 395.
10. RICE, J. R. and ROSENGREN, G. F., J. Mech. Phys. Solids, 16, 1968, 1.
11. DOI, H., FUJIWARA, Y. and MIYAKE, K., Trans. Met. Soc. AIME, 245, 1969, 1457.
12. LINDAU, L., Scand. J. Met., in print.
13. GURLAND, J., Trans. Met. Soc. AIME 227, 1963, 1146.

Table I Results of quantitative metallographic measurements and fracture toughness determination, the error in K_{IC} is the variance in a set of twelve measurements. E_c and σ_c are the elastic modulus and 0.2% compressive flow stress quoted by the manufacturer.

Alloy Number	f_β	d_α, μm	d_β, μm	K_{IC} MPam$^{1/2}$	E_c, MPa	σ_c, MPa
1	0.10	3.0	0.70	10.8 ± 0.3	6.3×10^5	3300
2	0.10	1.1	0.25	9.2 ± 0.4	6.3×10^5	4300
3	0.15	2.9	0.95	17.3 ± 0.9	5.9×10^5	2700
4	0.25	3.6	1.90	23.1 ± 2.4	5.4×10^5	2100
5	0.23	0.95	0.35	11.9 ± 2.4	5.4×10^5	2700

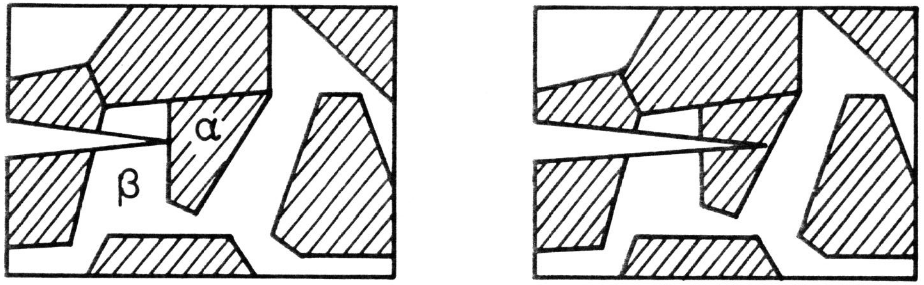

Figure 1 Two possible crack tip configurations.

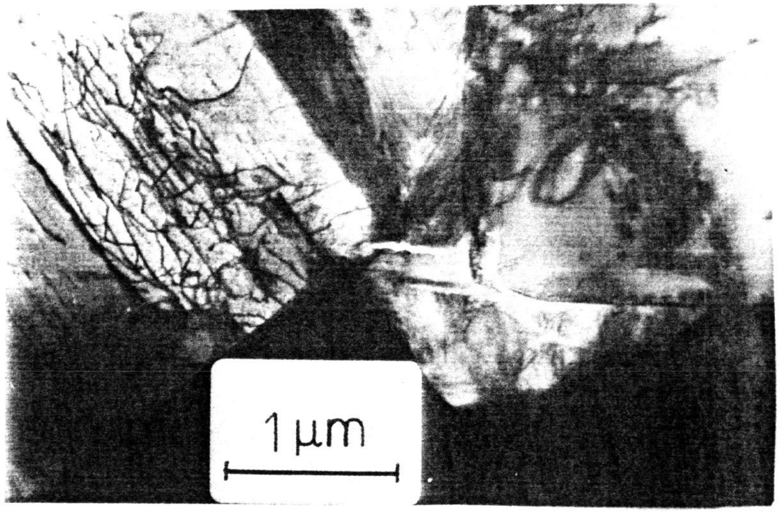

Figure 2 Transmission electron micrograph of a WC-15%Co alloy, deformed by a Vickers pyramid loaded with 50N. The micrograph was taken from an area located approximately 0.2 mm from the indentation edge. The features in the binder intersecting at the crack show a contrast which is similar to the c.p.h.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

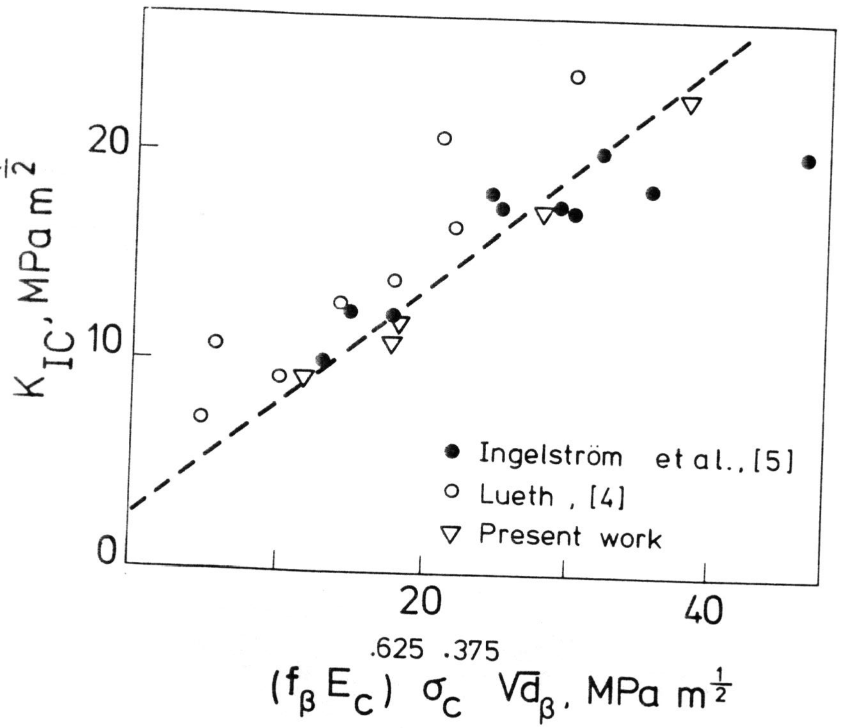

Figure 3 K_{IC} as a function of microstructural parameters.

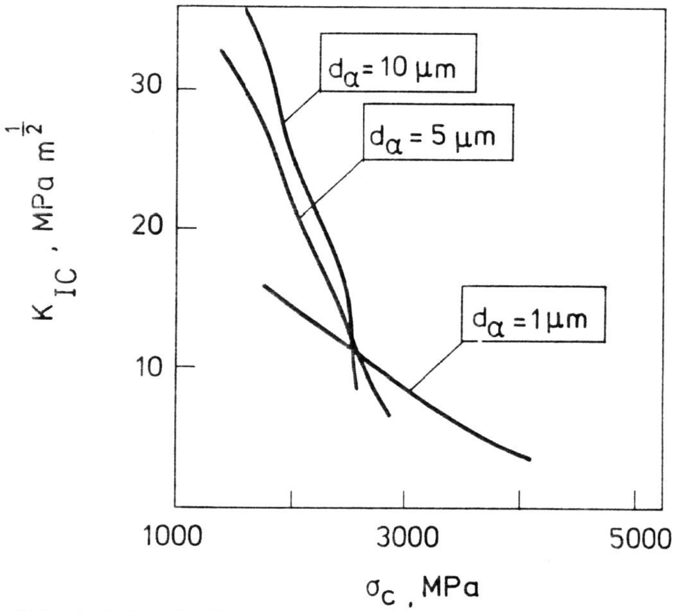

Figure 4   Calculated relation between K_{IC} and 0.2% flow stress for alloys with different carbide grain sizes.

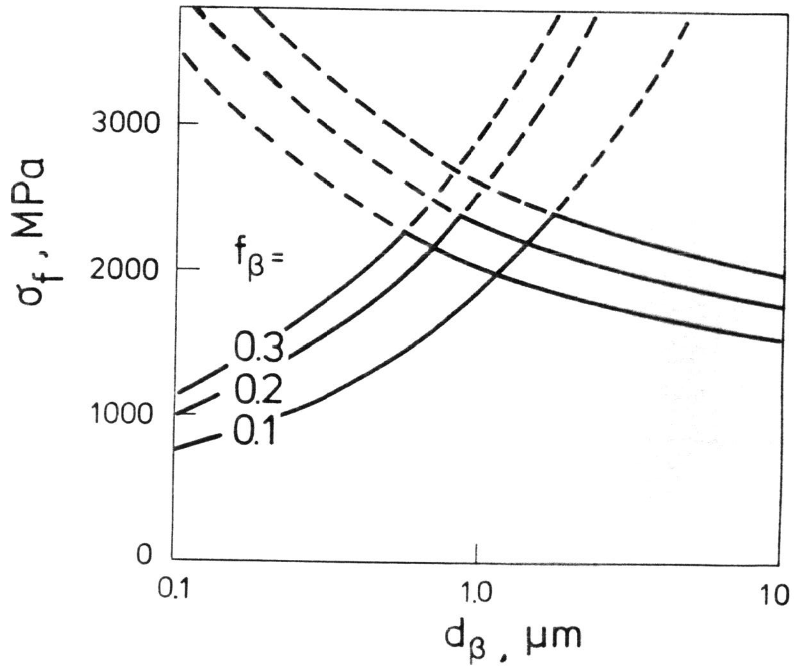

Figure 5   Estimated fracture stress as a function of d_β.