

Contribution to the discussion of brittle failure of cemented carbide tool materials

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CONTRIBUTION TO THE DISCUSSION OF
BRITTLE FAILURE OF CEMENTED CARBIDE TOOL MATERIALS

by

H.J.J. KALS

C.I.R.P.

Report presented at the meeting of the Subgroup "Toughness"
of the Technical Committee "C".

Paris, 31st January 1974.

Contribution to the discussion of

BRITTLE FAILURE OF CEMENTED CARBIDE TOOL MATERIALS

1. INTRODUCTION

Amongst the mechanisms which contribute to the deterioration of carbide cutting tools is the development and propagation of (micro) cracks leading to both small scale chipping and rupturing. This results in premature failure of the tool, even when tool loadings are comparatively low.

Large scale chipping or rupturing often is a result of high loadings causing stresses exceeding the rupture strength of the material and leading to catastrophic failure within a limited period of time. However, it has been shown by Ellis and Barrow ¹⁾ that even in the case of relative low feeds brittle failure of cutting tools by large scale chipping may occur when the ratio of the main cutting force to the feed force responds to certain conditions.

2. MECHANICAL PROPERTIES OF CEMENTED TUNGSTEN CARBIDES

2.1. GENERAL

It has been shown that the composition of cemented carbide materials, the distribution and the size of the grains and in particular the thickness of the matrix (cobalt) layers are of great importance with respect to both the strength and the hardness and subsequently to the brittle behaviour of these structured materials ²⁾.

It is known that when the cobalt content becomes smaller than about 6% the matrix becomes rather incomplete i.e. a plain matrix no longer exists. In this case the cobalt component is present in the form of thin layers covering the grains piled close together, whilst voids exist between them ³⁾.

¹⁾ J. Ellis and G. Barrow, Ann. C.I.R.P. vol. 21/1 (1972) 25.

²⁾ A.S. Tetelman and A.J. McEvily, Fracture of structural materials (1966).

³⁾ R. Arndt, Zeitschrift für Metallkunde Band 63 (1972) Heft 5, 274-280.

Plastic deformation of this kind of structure is only possible by

- relative displacement between the different grains, to a minor extent controlled by plastic deformation of the thin cobalt layers, but more substantially dependent on friction between the (covered) grains.
- plastic deformation and rupture of the WC-grains which are part of the various continuous grain-networks.

As the cobalt percentage is increased a plain matrix may exist. This will result in the whole structure more or less adapting the mechanical properties of the grains ⁴⁾.

Since the strength of the single grains far exceeds the strength of cobalt, relative small deformations will cause very high hydrostatic stresses in the cobalt layers, as a result of which

- the carbides tend to behave less brittle i.e. the conditions for plastic deformation are improved.
- high stress gradients will exist in the cobalt layers near the free surface or near the pores still present in the matrix.

With respect to the latter, it is to be expected that micro cracks are initiated by preference at the locations mentioned and that they occur at relative small loadings. This mechanism, however, seems (at least during compression) not to be of a dominating nature since Burbach's ⁵⁾ results show that the rupture strength by compression decreases with increasing cobalt content.

A further increase of the cobalt content reduces the influence of the mechanical properties of the carbides by increasing the thickness of the cobalt layers between the grains; in loaded condition the stresses will not exceed moderate values as the induced strains can be distributed more equally.

⁴⁾ J. Gurland, Trans. AIME 227 (1963) 1146.

⁵⁾ J. Burbach, Techn. Mitt. Krupp Forsch. Ber. Band 26. (1968) H2.

At high cobalt contents the mechanical properties of the cemented carbide material will approach those of pure cobalt.

2.2. THE STRENGTH

The influence of the various structural parameters on the strength of cemented carbides will not explicitly be mentioned here.

Mohr's theory predicts failure of "brittle" materials when

$$\sigma_1 - \left(\frac{\sigma_B}{\sigma_{-B}} \right) \sigma_3 \geq \sigma_B$$

or $\sigma_1 \geq \sigma_B$ when $\sigma_1 > \sigma_3 > 0$; $\sigma_2 = 0$

or $|\sigma_3| \geq \sigma_{-B}$ when $\sigma_3 < \sigma_1 < 0$; $\sigma_2 = 0$

In the case of carbide inserts this criterion has been applied to predict large scale chipping¹⁾.

One of the main problems using such an approach is the unreliability of strength properties of "brittle" materials.

Basically, rupturing can only take place as a result of tension stresses leading to local rupturing of the inter-atomic bonds. Due to the presence of micro-cracks which cause stress concentrations, the real rupture strength is considerably smaller than the theoretical value derived from the atomic forces:

$$\sigma_{cr} = \sqrt{\frac{2\gamma E}{a}}$$

in which a = the distance between the
atoms

γ = the energy necessary to
generate a new surface.

From this, Griffith derived a formula for the real rupture strength of completely amorph materials:

$$\sigma_{cr} = \sqrt{\frac{2\gamma E}{\pi c}}$$

in which $2c$ = the length of an initial micro-crack.

It appears that the rupture strength depends on the length of an initial crack. The presence of a number of micro-cracks with different length explains the scatter in experimental results on the rupture strength of "brittle" materials. This necessitates a statistical approach. In the case of ductile materials a substantial plastic deformation may precede brittle failure.

Orowan has shown that for materials which show some degree of plastic deformation before brittle failure, Griffith's equation may be modified into

$$\sigma_{cr} = \sqrt{\frac{E p}{c}}$$

(p = the work of plastic deformation in the vicinity of the tip of the growing crack)

which shows that the critical crack length is considerably higher ($p \gg \gamma$) than for completely amorph materials.

Irwin derived a stress intensity factor K which is applicable in the near vicinity of the tip of a crack

$$K = \sigma \sqrt{m c},$$

in which m is a geometrical factor

However, the quantities representing rupture strength, fatigue etc., do not only include propagation of the crack but also the initiation. Rupture mechanics do not deal with the latter phenomenon.

It is a fact that most "brittle" materials show a poor capability of plastic deformation combined with a low resistance against rupture in tension. Obviously those properties are interrelated and is brittle rupturing to a major extent controlled by the state of stress.

2.3. THE HARDNESS

The definition of hardness is the resistance against plastic deformation by indenting.

For one family of materials, an increase in hardness results in a decreasing capability of plastic deformation. Thus, harder materials behave more brittle when equal states of stress are induced.

Brittleness, however, is a phenomenon which seems not yet to be well understood. At least no proper definition of brittleness -or toughness- exists.

Few people tend to believe brittleness is a material property only. Most accept that brittleness represents a material behaviour depending upon the state of stress and the deformation rate.

With respect to this it can be mentioned that Rüdiger et al.⁶⁾ concluded that a definition of toughness of cemented carbides on the basis of one distinctive quantity only, must lead to problems.

In section 2.6. an attempt is made to give a systematical approach of the concept toughness, leading to the assumption of some characteristic parameters.

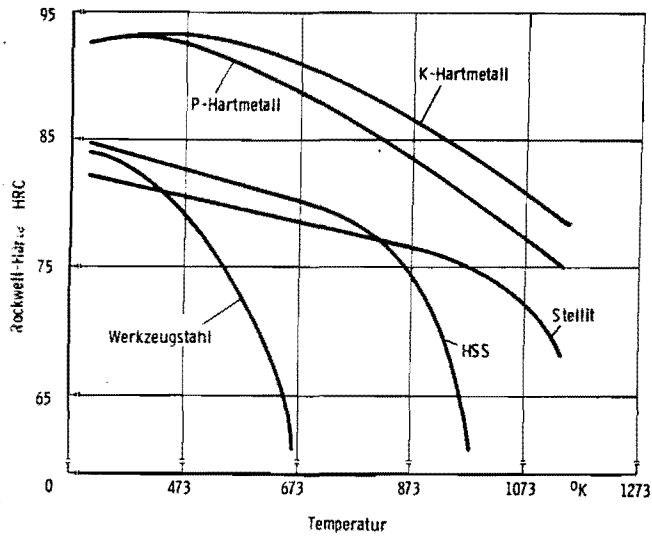
Orowan's formula suggest that there must be a relation between hardness and the sensitivity to brittle failure. The literature about rupture mechanics however seems not to deal with hardness explicitly.

The hardness of cemented carbides is a function of the cobalt content, whilst grain size is of minor importance.

The R_c - hardness as a function of temperature, as measured by Kreis⁷⁾, is shown in the next figure.

6) O. Rüdiger, G. Ostermann, H. Kolaska, Techn. Mitt. Krupp. Band 28 (1970) H2.

7) W. Kreis, Doctor's Thesis, Aachen (1973).

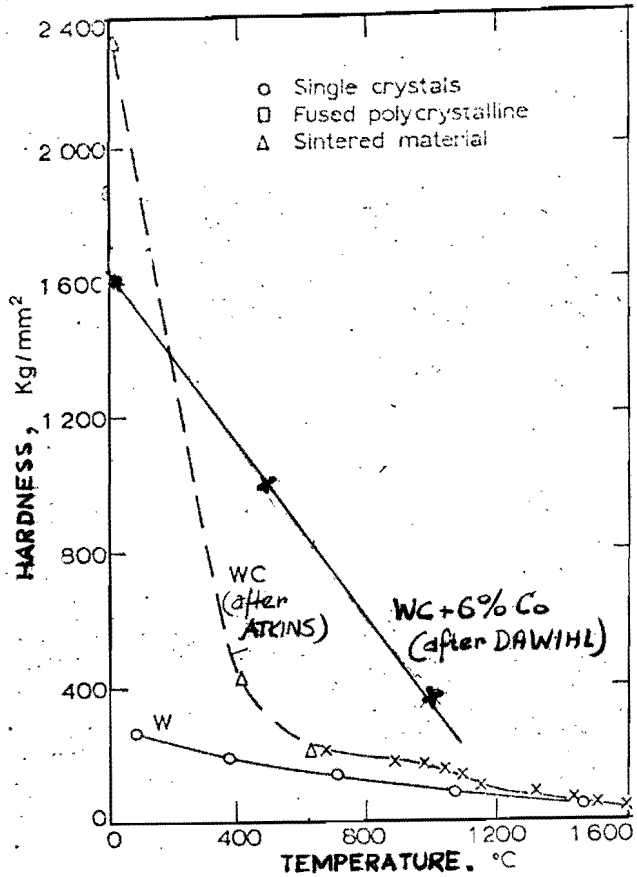


The next figure shows the Vickers hardness of W, WC and a cemented tungsten carbide with 6% Co as measured by Atkins and Tabor⁸⁾ and Dawihl⁹⁾ respectively. The results obtained by the latter concern a carbide grade resembling ISO - standard K01. .

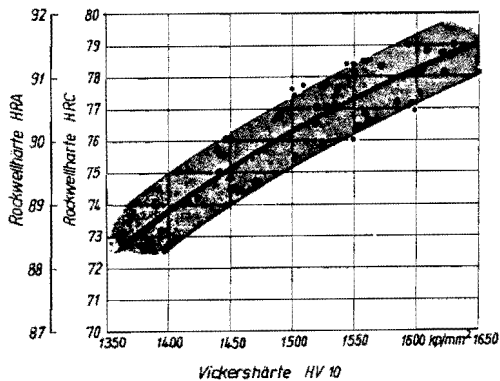
They show a remarkable drop in hardness when compared with Kreis's results.

⁸⁾ A.G. Atkins, D. Tabor, Proc. Roy. Soc. (1966) A 292, 441-459.

⁹⁾ W. Dawihl, E.A. Klingler, Technische Mitt. (Nov. 1968) Heft 11.



Although results reported by Lewald¹⁰⁾ suggest a relation between the Rockwell- and Vickershardness, when these methods are applied to cemented carbides it seems that the influence of scale-effects are not to be neglected. The results of Lewald are shown in the next figure.



¹⁰⁾ W. Lewald, Doctor's Thesis, Aachen (1962).

2.4. YOUNG'S MODULUS

The elastic behaviour of carbides has frequently been investigated. Usually the modulus of elasticity is derived from dynamic experiments by means of vibrating beams. It is observed that this quantity strongly depends on the percentage of cobalt. As a matter of fact a linear relation exists between the volume percentage Co and Young's modulus for $5 < \text{vol. \% Co} < 45$.

This range is for technical reasons the interesting one. The grain size does not much influence E.

The dependence of E on temperature is relatively weak: 1% per 100°C⁹⁾. With reference to the contents of section 2.6., the significance of Young's modulus is determined by the description of variations in the potential energy of the inter-atomic bonds as a result of different hydrostatic loadings.

Together with the coefficient of thermal expansion, Young's modulus determines the stress-sensitivity to thermal loading.

2.5. PLASTIC DEFORMATION PHENOMENA

Various investigators,⁵⁾³⁾¹¹⁾ working independently from each other, have come to the conclusion that the carbide grains take part in the process of plastic deformation of cemented carbide materials. From their experiences they deduce that, at least at compressive loading, the mechanical behaviour of carbides is merely controlled by existing continuous carbide-networks, even when the grains are separated by thin cobalt layers. It is stated by Arndt³⁾ that in the case of WC - Co structures the network-model is valid as long as the percentage of cobalt is smaller than 35% (= 50 Vol.%).

It is also observed by Arndt that the resultant deformation is composed of at least the next three mechanisms

- shearing of carbide grains
- sliding between different grains
- transition of the cobalt matrix from b.f.c. into a hexagonal configuration.

¹¹⁾ V.A. Ivensen e.o., Soc. Powder Metallurgy and Metal Ceramics 19(1964)300.

Moreover it is believed that shearing of the cobalt matrix takes place but that observation of shear lines is made impossible by the transition. The initiation of plastic deformation is determined by those stresses that are able to deform the network by sliding and by plastic deformation of the grains.

Burbach also noticed that during compression the plastic deformation which precedes rupture is distributed among both the carbide grains and the cobalt layers. The measure in which the grains take part in plastic deformation depends upon the cobalt content and decreases with increasing cobalt percentage.

As all the results are derived from compression tests, not much is known about the plastic behaviour of cemented carbides during tension. Compression tests tend to induce some hydrostatic pressure (plain strain) and consequently during tension tests the grains may show a more distinct preference to rupturing instead of deforming plastically (see also section 2.6.). Secondly the adhesion strength between both the carbides and the cobalt layers may become more decisive. Here, however, one should remind Arndt's results where during compression tests the initiation of plastic deformation showed to be predominated by the properties of the carbide "network" up to cobalt contents of 35%³⁾.

The third and probably most important difference is that contrary to the behaviour during negative stresses, during tension the inhomogeneous nature of the structure becomes most decisive.

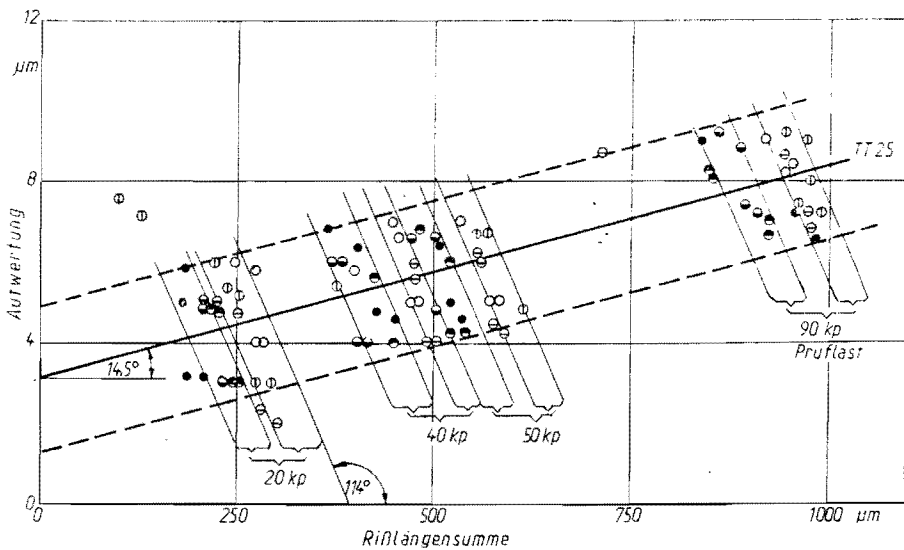
Impurities, pores, etc., will cause stress concentrations leading to an apparent reduced rupture strength. At this place the theory of rupture mechanics could provide us with a stress concentration coefficient.

(This coefficient cannot be represented by the ratio $\frac{\sigma_B}{\sigma_{-B}}$, since the state of stress during tension tests is different from that during compressive tests).

2.6. TOUGHNESS

Apart from the resistance to wear, toughness is the most important quantity when cemented carbide tool materials are involved. In contrast with this there is no unanimately accepted quantity to represent toughness. The two most current criteria are firstly the maximum plastic strain up to failure (δ_{\max}) and secondly the total work absorbed (A_t) till failure.

A first remark with respect to the latter is that it is disputable whether the energy which is consumed by the initiation and propagation of cracks has to be taken into account or not. During compressive loadings, the energy consumed by the formation of cracks may be many times greater than will be the case under tensile forces. However, it is to be expected that for both processes the ratio of the consumed energy by rupturing to the work for plastic deformation $A_r/A_{pl} \ll 1$. Accordingly the calculation of A_{pl} would satisfy. Moreover, it appears to the author that for technical reasons the application limit is better marked by crack initiation than by failure. The same can be said for δ_{max} . Rüdiger⁶⁾ reported from experiments where the total length of cracks caused by Vickers-indents are measured as a function of the height of the bulge which is formed in the near vicinity of the indent. The height of the bulge is considered to be a measure for the work ($A_{pl} + A_r$). The results for various loadings are represented in a graph (see next figure) and the intercept caused by the averaging curve represents the plastic work up to crack initiation.



This approach has originally been reported by Dawihl in 1940. Houdek, Dlouhy and later Palmqvist¹²⁾ developed methods for the determination of the toughness of cemented carbide materials based on this idea. Houdek et al.

¹²⁾ S. Palmqvist, Arch. Eisenhüttenwes. 33 (1962) 629.

applied the permissible load and the related diameter of the dent for crack initiation as a toughness criterion. The latter considered the energy consumed by the formation of cracks as a measure for toughness. The author prefers Rüdiger's method, but emphasizes Rüdiger's own remark that the height of the indent is not an adequate measure for the plastic work. A better method for the calculation of the plastic work could possibly be derived from the energy balance. Another objection to the method is that the total plastic work as a quantity on its own will not determine toughness sufficiently since in various cases the volume of involved material is different. Subsequently the (average) plastic work per unit volume of the material involved may be a more adequate quantity. This, however, has to be investigated. It is a fair possibility that a proper relation exists between hardness and the specific work up to crack initiation.

The problem is still left which definition of toughness suits best.

As to the two definitions mentioned (i.e. δ_{\max} and $A = A_{el} + A_{pl}$), for each of them preferential application areas can be indicated.

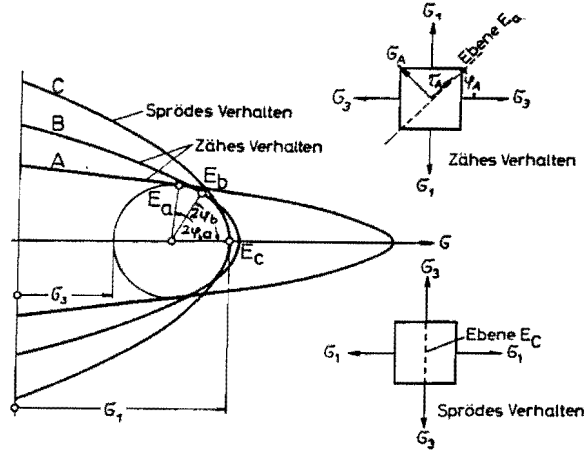
In the case of forced low speed deformations which can be caused by e.g. by slow thermal expansion, δ_{\max} - and better still $(\delta_{el} + \delta_{\max})$ - is probably a very suitable toughness criterion whilst the application of A as a criterion will not satisfy.

On the otherhand, in the case of impact loading or high deformation rate, the work A is decisive for the behaviour of the material whilst $(\delta_{el} + \delta_{\max})$ is not suitable. One of the reasons for this is the influence of the strain rate (Burbach noticed that there is a clear influence of the strain rate).

Concluding, one can say that the type of loading in the time domain plays a major role in the choice of the toughness criterion.

It has been mentioned before that brittleness - or the inverse, toughness - depends on the state of stress and the deformation rate. The dependence on the state of stress can be demonstrated with the aid of the next figure (after Leon; the figure is borrowed from Primus¹³⁾).

¹³⁾ I.F. Primus, Doctor's Thesis, Aachen (1969).



Looking at this figure some interesting thoughts arise.

When applying a load to a body the induced stresses will adapt to the load till equilibrium is reached. The resulting state of stress (determined by the type of load and the conditions mentioned) can be characterized by a hydrostatic pressure p and

$$\begin{aligned} \sigma_1 &= k_1 \tau_A & \sigma_1 &> \sigma_2 > \sigma_3 \\ \sigma_A &= k_2 \tau_A \\ \sigma_3 &= k_3 \tau_A & k_3 &< k_2 < k_1 \end{aligned}$$

where k_1 , k_2 and k_3 are constants.

With increasing load, the stresses rise in proportion until either plastic flow or rupture occurs.

Let σ_{BE} be the effective rupture strength and τ_f the flow shear stress, then complete brittle rupture will take place when

$$\frac{\tau_A}{\tau_f} < \frac{\sigma_1}{\sigma_{BE}} = 1$$

In the opposite case plastic deformation will take place first. The conditions for this case are :

$$\frac{\sigma_1}{\sigma_{BE}} < \frac{\tau_A}{\tau_f} = 1$$

In the degree of plastic deformation which can now be achieved is depending on two mechanisms :

Firstly, with continuing deformation the flow shear stress will increase as a result of strainhardening.

Secondly, micro-cracks that are initiated in the shear zone show the most unfavourable position with respect to the normal stress, σ_A , on the shear plane. This, combined with impurities accumulating at certain places on the shear plane, will lead to a decreasing effective rupture strength in the direction of $\sigma_A : \sigma_{BEA}$.

The rupture strength in the l-direction (see Figure) will not substantially be affected by either of the two phenomena.

Subsequently with continuing deformation two types of rupture may occur.

1. When $\frac{\sigma_A}{\sigma_{BEA}} < \frac{\sigma_1}{\sigma_{BE}} = 1$, rupture will take place

in a plane perpendicular to the highest main stress.

2. When $\frac{\sigma_1}{\sigma_{BE}} < \frac{\sigma_A}{\sigma_{BEA}} = 1$, rupture will take place

in the shear plane.

The rupture strength will not be influenced much by the strain hardening effect. It is believed, however, that the effective rupture strength is a function of the hydrostatic stress p .

As plastic deformation is completely described by deviatoric stresses, the hydrostatic stress controls the energy conditions for failure by changing the equilibrium position of the atoms.

The presence of inhomogenities and micro-cracks in the original material will also affect the effective rupture strength. The application of stress concentration factors can remedy this.

This way of thinking leads to the idea that the main parameters for toughness are :

1. $\frac{\tau_f}{\sigma_{BE}}$ and $\frac{\sigma_{BE}}{\sigma_{BEA}}$ (both are functions of the hydrostatic stress p and the deformation δ)

2. k_1 and k_2

3. ADDENDUM

Other properties which are important with respect to tool materials such as mechanical and thermal fatigue have not yet been dealt with. Moreover, important with respect to fracture analysis are the methods to derive the stresses in the tool material caused by mechanical and thermal loads.

Although there are many other problems to be solved in this field, the author feels that the investigations into brittle fracturing of cemented carbides is hampered mostly by an insufficient knowledge of the mechanical properties of these materials.

It is the hope of the author that this report will start a discussion on future work in the field of cemented-carbide tool materials.

Eindhoven, 28th January 1974

H.J.J. Kals

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