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Influence of notch radius and R-curve behaviour on the fracture toughness evaluation of WC–Co cemented carbides

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ABSTRACT

In this work the apparent fracture toughness of different grades of WC–Co cemented carbides has been evaluated by means of the single edge V-notched beam (SEVNB) method. The results are analysed, as a function of both the binder mean free path and the maximum steady-state crack shielding value, following a theoretical approach which assumes the existence of a small crack at the notch tip. The experimental findings indicate that larger defects at the notch tip and/or existence of R-curve behaviour imply higher (i.e. less restricted) values of the minimum notch radius required in SEVNB specimens for a reliable assessment of the fracture toughness of cemented carbides.

1. Introduction

The correct evaluation of the fracture toughness, K_{Ic}, of brittle materials has an especial significance in common applications which involve impact and/or monotonic loading. In this regard, for many applications of WC–Co cemented carbides (materials usually referred to as hardmetals), such as metalforming tools (e.g. press-forming dies) or structural components (e.g. seals, jaws, pins, etc.), toughness represents a critical parameter for material selection and design. This intrinsic property has historically adopted different meanings, especially within the field of hardmetals, such as transverse rupture strength (TRS), energy absorption as related to (dynamic) impact testing, Palmqvist/Vickers indentation toughness and fracture toughness as related to linear elastic fracture mechanics (LEFM). Nevertheless, specific limitations associated with all the conventional techniques employed for the evaluation of the fracture toughness of these materials have been reported since early investigations [1].

The fact that brittle fracture of cemented carbides mostly initiates from pre-existing flaws has supported design methodologies based on LEFM. As a consequence, its use as a tool for the fracture analysis of these materials has been well established [2–4], and since its introduction in the sixties some of the previous limitations for K_{Ic} evaluation have been reduced, particularly regarding the empirical nature of this parameter (different type, size and distribution of natural flaws). Some pioneer works in the field of fracture toughness assessment in cemented carbides are those from Kenny [1], Chermant et al. [5], Exner et al. [6], Lueth [7] and Inglesträm and Nordberg [8].

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Nomenclature

a_0	initial length of the notch
$2a_c$	critical natural flaw sizes
CNB	Chevron notched beam
CT	compact tension
d_{WC}	carbide grain size
DCB	double cantilever beam
DT	double torsion
$F_{bend}(a \text{ or } a_0/W)$	shape function for a cracked geometry
IM	indentation microfracture
IF	indentation flexure
K	stress intensity factor for a cracked specimen
K^*	stress intensity factor for specimens with sharp pre-cracks
$\hat{K}$	stress intensity factor for specimens with V-notch
K_{appl}	applied stress intensity factor
K_{sh}	shielding stress intensity factor
$K_{sh,max}$	maximum shielding stress intensity factor
K_t	stress intensity factor for crack initiation
K_{total}	total stress intensity factor
K_{Ic}	fracture toughness
K_{I0}	intrinsic toughness
K_R	R-curve behaviour
l	final crack length
l_0	initial crack length
LEFM	linear elastic fracture mechanics
R	notch root radius
$R_{c,theor}$	theoretical critical notch root radius
R_{exp}	experimental notch root radius
SCF	surface crack in flexure
SENB	single edge notched beam
SEPB	single edge pre-cracked beam
SEVNB	single edge V-notched beam
V_{Co}	volume fraction of Co
Y	shape function for natural flaws
σ_{bend}	failure stress under flexure
σ_R	experimental flexural strength
β, λ	constants associated with the microstructural parameters and shielding
λ_{Co}	binder mean free path
α, χ	proportionality factors
$\Delta a = l - l_0$	subcritical crack extension
δ	variation of K_{total}/K_{I0} as a function of Δa

The effective application of LEFM for the correct evaluation of the fracture toughness of brittle materials has been pursued introducing several methodologies which involve different pre-cracking techniques, specimen geometries and crack configurations: single edge pre-cracked beam (SEPB) [9], single edge notched beam (SENB) [10–13], Chevron notched beam (CNB) [14,15], surface crack in flexure (SCF) [16], single edge V-notched beam (SEVNB) [17,18], indentation microfracture (IM) [19], indentation flexure (IF) [20], double cantilever beam (DCB) [21], compact tension (CT) [8] and double torsion (DT) [22]. Among all of them, IM has been widely employed for the fracture toughness evaluation in brittle materials due to its ease and cost efficiency [19]. From this viewpoint, although much effort has been put in the past and at present in order to validate its use for cemented carbides [23], it exhibits inherent limitations such as (1) uncertainties in the residual stress field developed by the indentation, (2) wide range of parameter calibrations [23], (3) crack size sensitivity as a function of surface preparation [24], and (4) chipping effects inherent to certain materials and/or indentation loads, among others, that may lead to an overestimation of the toughness values which in some cases turns into a rejection of the method, e.g. for relatively tough materials [25].

In this regard, a more reliable methodology is that which requires the presence of a fine crack through the specimen section, initiated from a notch [9,10,12–14]. Thus, the correct evaluation of the fracture toughness is based on the suitability of the procedure for introducing sharp pre-cracks in a specimen, free of residual stresses, and testing under a completely defined stress state. Evidently, this turns into a difficult task especially for the case of brittle materials such as cemented carbides. In these cases, the stress intensity factor, K^*, is given by the following equation [18,26]:

$$K^* = \sigma_{\mathrm{bend}} \sqrt{\pi(a_0 + l)} F_{\mathrm{bend}} \left(\frac{a}{W} \right), \tag{1}$$

where σ_{bend} is the specimen failure stress under flexure, a_0 is the initial length of the notch, l is the crack generated in front of the notch tip and $F_{\mathrm{bend}}(a/W)$ is a factor associated with the specimen geometry and the loading conditions for an edge crack of length $a_0 + l$ within a sample of height W. The validity of this method requires the length of the crack in front of the notch, l, to be larger than the notch radius, R, being $l/R = 1.5$ a lower bound estimation [27]. However, the techniques employed to generate such pre-crack are often expensive and require the expertise of a specialist. This turns in most cases into a non-reproducible result, fact also associated with the difficulty in the precise measurement of the crack length.

In an attempt to overcome the problem of generating and/or detecting the pre-crack, the (apparent) stress intensity factor $\hat{K}$ has been evaluated in prismatic specimens with a V-notch (SEVNB) using the following equation:

$$\hat{K} = \sigma_{\mathrm{bend}} \sqrt{\pi a_0} F_{\mathrm{bend}} \left(\frac{a_0}{W} \right), \tag{2}$$

where the total crack length is assumed as the length of the notch, a_0. Nonetheless, it is well known that the SEVNB method may not be a reliable approach to evaluate the absolute fracture toughness due to the experimental limitations inherent to the technique itself [28,29], which may be even more significant for brittle materials with a fine microstructure or under the presence of crack growth resistance (R-curve) behaviour before final fracture, such as the case of cemented carbides [30].

In this work, the fracture toughness of several WC–Co cemented carbides grades are evaluated based on experimental results and different theoretical models in order to establish the suitability of the SEVNB method for the correct assessment of the fracture toughness of these materials, taking into account the influence of both notch radius and R-curve (crack shielding) effects.

2. Influence of the notch radius and R-curve on toughness evaluation: theoretical considerations

The effective application of LEFM for the correct evaluation of the fracture toughness of a brittle material requires the presence of a crack. Hence, let us consider the case of a small crack in front of a sharp notch machined in the specimen. This small crack, which is considered to extend through the entire thickness of the specimen, i.e. an edge crack, can be thought as developed from flaws in the material (natural) and/or damage-induced defects during the machining procedures. These flaws are assumed as randomly distributed along the notch tip [28,29].

Fett and Munz have proposed an equation to describe the stress intensity factor, K, of a crack in front of the notch tip of a SEVNB specimen, which is applicable for $l \ll a_0$ [27,31], as follows:

$$K \approx \hat{K} \tanh \left(2.243 \sqrt{\frac{l}{R}} \right), \tag{3}$$

where $\hat{K}$ is given by Eq. (2) and l/R is the ratio between crack length and notch radius. The ratio between K and $\hat{K}$ is represented in a two-dimensional plot in Fig. 1 as a function of crack length l and notch radius R. It can be observed that the effect of the notch radius is no longer significant when $l/R > 1.5$, as demonstrated by other authors [27]. In such a case, the fracture toughness may be correctly assessed.

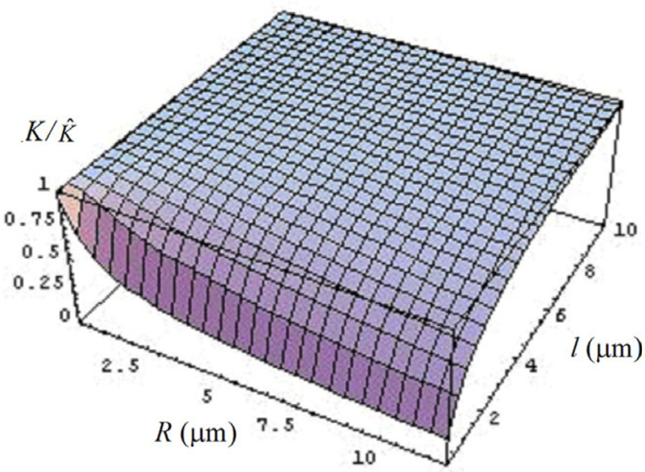

Fig. 1. Variation of $K/\hat{K}$ with the crack length l and notch radius R.

However, in the case of materials which show R-curve behaviour, the total stress intensity factor, K_{total}, is given by the superposition of the applied stress intensity factor, K_{appl}, and that associated with the shielding effect, K_{sh}. The magnitude of the shielding intensity factor varies with the crack length until reaching a maximum saturation value $K_{sh,max}$, and can be described for most materials by the following expressions:

$$K_{sh} = K_{sh,max}\left[1 - \exp\left(-\frac{\Delta a}{\beta}\right)\right], \tag{4}$$

$$K_{sh} = K_{sh,max}\left[1 - \exp\left(-\sqrt{\frac{\Delta a}{\lambda}}\right)\right], \tag{5}$$

where $\Delta a = l - l_0$ is the crack extension between the final (l) and initial (l_0) crack length, and β and λ are constants associated with the shielding of a given material. Eq. (4) describes the case of materials with an initial linear R-curve, whereas Eq. (5) represents the one of materials with steep R-curve behaviour [32].

Considering the R-curve effect in a material whose shielding may be described by Eq. (4), the K_{total} which is associated with crack growth can be obtained by adding the intrinsic K_{sh} to the externally applied stress intensity factor K_{appl} (equal to K, as given by Eq. (3)). Thus adding Eqs. (3) and (4) and normalising by the material intrinsic toughness K_{I0}, the relation between K_{total} and K_{I0} can be expressed as

$$\frac{K_{total}}{K_{I0}} = \alpha \cdot \tanh\left[2.243\sqrt{\frac{l}{R}}\right] - \chi \cdot \left[1 - \exp\left(-\frac{\frac{l-l_0}{R}}{\frac{\beta}{R}}\right)\right], \tag{6}$$

where both α and χ are the proportionality factors corresponding to $\hat{K}$ (stress intensity factor associated with a crack of length a_0) and $K_{sh,max}$ (maximum shielding), both with respect to K_{I0}, respectively.

It can be inferred from Eq. (6) that the evaluation of the fracture toughness by the SEVNB technique in a material with crack growth resistance behaviour will be influenced by several parameters: (1) the initial crack length, l_0, (2) the notch radius, R, and (3) the shielding nature and characteristics of the material. In Fig. 2 the normalised total stress intensity factor is represented as a function of the crack length/notch radius ratio (l/R) for two materials with different maximum shielding levels, i.e. (a) $\chi = 1$ (Fig. 2a) and (b) $\chi = 2$ (Fig. 2b). In order to better identify the different stadiums of crack growth, the normalised initial crack length l_0/R is assumed to be 0.02. The constant β is taken as 0.3 μm, based on practical cases. Two regions of stable/unstable crack growth can be observed in both cases. For small values of $\hat{K}$ no crack growth is expected. When the apparent stress intensity factor $\hat{K}$ increases, the ratio K_{total}/K_{I0} may be greater than 1 and thus the crack starts growing. The condition of stable/unstable crack growth is given by the variation of Eq. (6) as a function of crack length increment (Δa), represented by $\delta = d(K_{total}/K_{I0})/da$. It can be inferred from Fig. 2a and b that for a certain $\hat{K}$ value, a critical condition may be found where Eq. (6) is tangent to $K_{total}/K_{I0} = 1$. Finally, for large values of $\hat{K}$, the initial stable growth will be always followed by unstable growth ($\delta > 0$) leading to the material failure. Comparing the referred figures it can also be discerned that the stable crack growth is proportional to the shielding level (R-curve) given by the parameter χ.

The influence of the constants β and α on the stable growth of the crack is presented in Figs. 3 and 4, for a given l_0/R ratio of 0.08, estimated for small initial cracks as the size of a fine microstructure and a corresponding feasible notch radius of 10 μm. Three different shielding levels, $\chi = 1$, 1.5 and 2, based on interesting practical cases are assessed. By analysing the referred figures various behaviours, independent of the intrinsic toughness of the material, K_{I0}, may be indicated: (1) the amount of stable crack growth increases as the constant β takes higher values, regardless of the shielding capability of the material; (2) the shielding effect is more significant for high values of β; (3) for a given K_{I0}, the value of $\hat{K}$ decreases

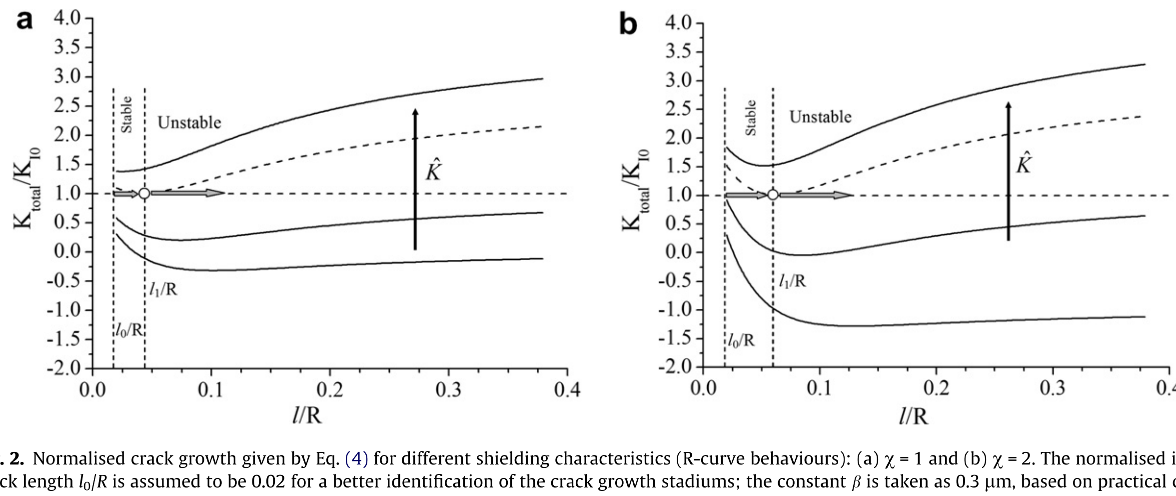

Fig. 2. Normalised crack growth given by Eq. (4) for different shielding characteristics (R-curve behaviours): (a) $\chi = 1$ and (b) $\chi = 2$. The normalised initial crack length l_0/R is assumed to be 0.02 for a better identification of the crack growth stadiums; the constant β is taken as 0.3 μm, based on practical cases.

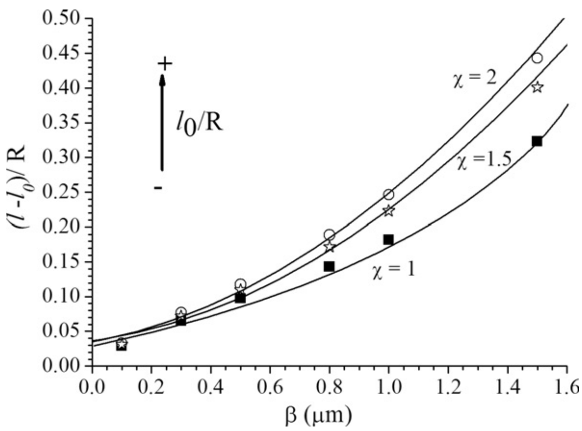

Fig. 3. Influence of the parameter β on the stable crack growth, independent of K_{I0}, for $l_0/R = 0.08$ and $\chi = 1$, 1.5 and 2. An increase of the l_0/R ratio, i.e. smaller R, would shift the corresponding curves upwards, leading to an increase of the stable crack growth capability.

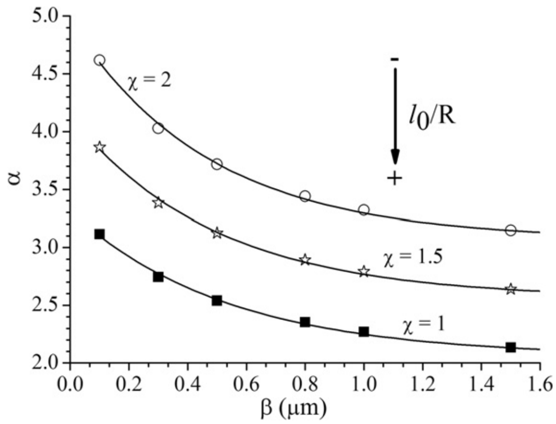

Fig. 4. Relation between the parameters β and $\alpha\,\hat{K}/K_{I0}$ *independent of* K_{I0}, for $l_0/R = 0.08$ and $\chi = 1$, 1.5 and 2. An increase of the l_0/R ratio, i.e. smaller R, would shift the curves downwards, resulting in a smaller $\hat{K}$.

as the value of β increases for every shielding level; the same tendance of $\hat{K}$ can be appreciated for the different shielding cases independent of β.

A general analysis attempting to account for the influence of the different geometrical and material parameters on the evaluation of the total stress intensity factor in a notched specimen has been presented. Below, a similar approach is implemented for analysing toughness evaluation of WC–Co cemented carbides.

3. Crack growth behaviour in notched WC–Co materials: experimental results and discussion

There exists extensive information about the relationship between microstructure and resistance to crack propagation for high-toughness ceramics and composites. From this perspective, cemented carbides have usually been considered as an academic example of toughening of ceramics by ductile ligament reinforcements. The main features which define the microstructure of WC–Co cemented carbides are the volume fraction and physical dimensions of each phase, i.e. the mean carbide grain size (d_{WC}) and the binder mean free path (λ_{Co}). Under the consideration of potential failure origins as small processing flaws without any initial ductile ligament zone (e.g. pores, large carbides or binderless carbide agglomerates), it is clear that any stable crack growth (i.e. before failure) would imply variable crack growth resistance (R-curve) characteristics. In this regard, the most important parameter to increase the R-curve behaviour of hardmetals is the effective volume of reinforcing metallic ligaments, at the crack wake behind the notch tip, which shields the propagating crack.

Concerning R-curves, they are experimentally complicated to determine as well as time consuming. However, a qualitative description of such behaviour, suitable enough for implementation in a rupture model, may be attained from basic ideas relating toughening mechanisms, fracture toughness assessed for "large" cracks (i.e. once crack-tip shielding has reached the plateau value) and the intrinsic fracture resistance of the material. In this work three cemented carbide grades named as 16F, 16M and 27C, with different carbide grain size and metallic binder volume fraction, have been investigated. Their

Table 1
Microstructural parameters, maximum shielding, flexural strength, experimental flaw sizes and critical notch radius for the three materials investigated

WC–Co	d_{WC} (μm)	V_{Co} (%)	λ_{Co} (μm)	σ_R (MPa)	$-K_{sh,max}$[a] (MPa m$^{1/2}$)	$(2a)_{exp}$ (μm)	$R_{c,theor}$ (μm)
16F	0.50	16.3	0.25	2742 ± 130	9.2 ± 0.6	8–14	3.1
16M	1.06	16.4	0.30	1813 ± 134	10.5 ± 0.8	14–34	3.7
27C	1.66	27.4	0.76	2629 ± 241	14.7 ± 1.0	30–42	9.5

[a] Estimated values from the experimental results obtained with SENB specimens, pre-cracked under cyclic compression [30].

microstructural parameters are given in Table 1. The following fundamental considerations regarding the crack growth resistance of these materials may be summarised as follows: (1) toughness enhancement is mostly given by the energy expended in the constrained plastic stretching of the cobalt binder ligaments, over a region behind the crack-tip of about 4 or 5 d_{WC} in length [33]; (2) toughness increases with crack extension up to a maximum steady-state level corresponding to the bridging length where the ligament zone is totally developed, i.e. K_{Ic} for large cracks which corresponds to $K_{sh,max}$ (see Table 1); (3) the microstructural parameters of the studied materials, i.e. d_{WC} and λ_{Co}, are relatively fine; and (4) the critical crack-tip stress intensity factor required for crack initiation, K_t, includes not only the intrinsic toughness of the WC matrix but also additional toughening mechanisms, particularly crack deflection, thus resulting in an effective value of about 7 MPa m$^{1/2}$ [33]. Accordingly, although many expressions may be found in the open literature for fitting R-curve behaviour (K_R), the very-steep character evidenced for hardmetals, which may be initially described by a linear function, points out the suitability of an exponential-like expression as given by Eq. (4). In such a case, the value of the β constant is assumed as two times the binder mean free path, i.e. $2\lambda_{Co}$, associated with the size of zone with Co ligaments, whereas the maximum shielding can be estimated on the basis of the fracture toughness experimentally assessed on notched specimens pre-cracked under cyclic compression tests [30] (Table 1). Hence, the modelled R-curve behaviour for the three cemented carbides grades here investigated is presented in Fig. 5. The maximum shielding corresponding to the tangent condition in Fig. 5 may be obtained from the experimental flexural strength results (4-point bending, with inner and outer spans of 20 and 40 mm, respectively, on bar-shaped specimens of geometry 45 mm × 4 mm × 3 mm), σ_R, together with the critical natural flaw sizes, $2a_c$ (Table 1), using a simple fracture mechanics model where natural flaws are taken as an embedded circular crack with a corresponding $Y = 2/\pi$ [4]. Within this framework, it can be indicated that the R-curve behaviour and the fracture toughness values associated with small cracks (natural flaws) vary with λ_{Co}. Moreover, the fracture toughness results estimated from natural flaws are found to be, in all cases, lower than those experimentally obtained with large cracks. This observation manifests the role of the crack growth resistance mechanisms present in these materials, which should be taken into account in quality control assessments of hardmetal tools and/or components.

Figs. 6 and 7 show the influence of l_0/R on the amount of stable crack growth and the α value (where K_{I0} is the fracture toughness of the carbide phase [34]), respectively. From the referred figures the following aspects may be highlighted: (1) stable crack growth is directly proportional to l_0/R and effective volume of binder ligaments (R-curve behaviour), independent of the shielding level and the l_0/R ratio, respectively; (2) $\hat{K}$, that is, the fracture toughness that would be measured without considering the existence of small cracks in front of the notch, diminishes with l_0/R regardless of shielding effects; (3) $\hat{K}$ increases with the binder mean free path of the ligament phase for a given l_0/R; and very important (4) no significant variations of the corresponding α are evidenced, independent of shielding effects, for l_0/R values greater than 0.08. Within this context, and considering that the size of the crack in front of the notch tip is proportional to the microstructural unit of these materials (λ_{Co}), a critical notch radius $R_{c,theor}$ can be estimated such that, for any R values below it, the fracture toughness assessed remains constant and corresponds to the level determined on notched and pre-cracked specimens (Table 1).

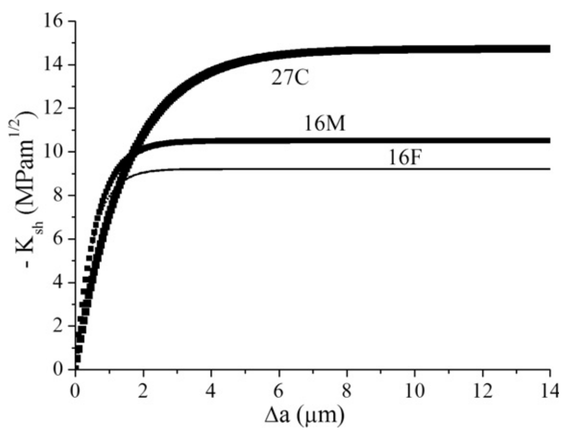

Fig. 5. Variation of the shielding factor, K_{sh}, for the three WC–Co investigated.

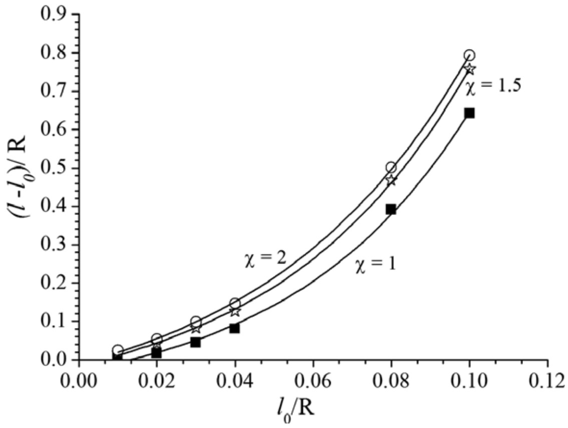

Fig. 6. Influence of l_0/R on the stable crack growth of the materials with different shielding characteristics.

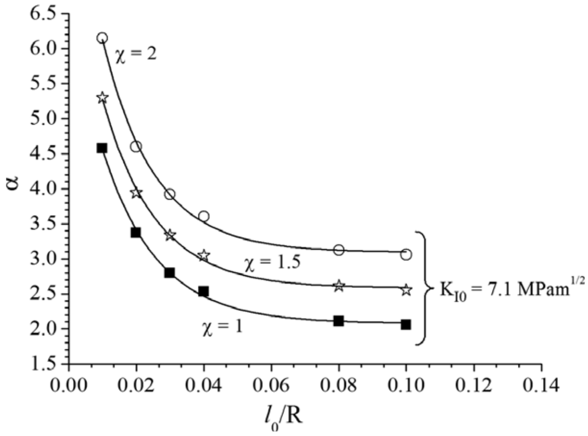

Fig. 7. Influence of l_0/R in the parameter α for materials with different shielding characteristics. No significant variations of α are observed for $l_0/R > 0.08$.

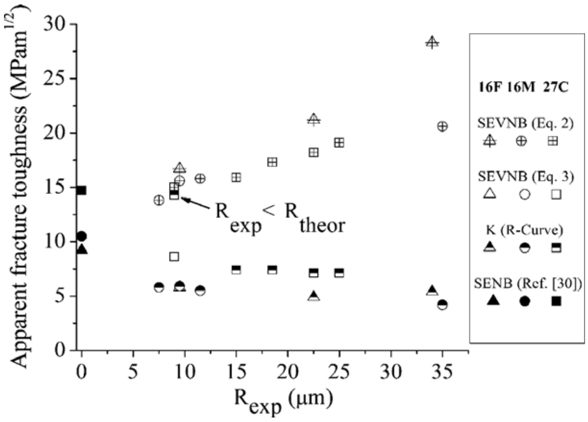

Fig. 8. Comparison between the apparent fracture toughness values obtained from notches and from large cracks.

In Fig. 8 the values of fracture toughness determined on notched specimens following different criteria are presented. They are compared with the experimental results measured in specimens with large cracks, taken here as reference levels [34]. Analysing the results evaluated for different notch radius (considering cracks of length a_0, as in Eq. (2)), it can be observed that the SEVNB toughness values are always higher than the reference ones, being this overestimation directly and indirectly related to the notch radius and the binder mean free path of the ligament phase, respectively. On the other hand, if fracture toughness is evaluated just assuming a fine crack of length l_0 in front of the notch tip, of the order of the microstructural size (as given by Eq. (4) [27]), the attained values are then even lower than the experimentally determined for long cracks, i.e. in this case the subcritical crack growth experienced by the WC–Co, and the implicit R-curve behaviour, is not taken into account. Thus, an alternative analysis option is proposed by considering the amount of stable crack growth experienced by the pre-existing crack (in from of the notch tip) until the tangent condition, $l - l_0$ (see Fig. 2). As evidenced in Fig. 8, and given that the notch radius is lower than the critical radius theoretically estimated (this condition being only satisfied, within the range of experimental variables here studied, for the 27C grade), toughness values assessed on notched specimens are then in satisfactory agreement with those obtained from specimens with large cracks. Similar findings would be expected for the other two hardmetal grades if specimens with machined notches of radius lower than 3 μm were available; however, this has been an experimental constraint impossible to overcome in this study.

4. Conclusions

The effective implementation of the SEVNB method for fracture toughness evaluation seems more feasible and costly effective for materials that experience R-curve behaviour and/or contain defects at the notch tip. This becomes more relevant and significant when the shielding effect and/or size of defects increase.

For the particular case of WC–Co, the critical values of l/R above which the fracture toughness remains constant is directly associated with the effective volume of binder ligaments reinforcing the composite (shielding effect). The experimental findings indicate that larger defects at the notch tip and/or existence of R-curve behaviour imply higher (i.e. less restricted) values of the minimum notch radius required in SEVNB specimens for a reliable assessment of the fracture toughness of cemented carbides. As a consequence, a systematic evaluation of the influence of the ratio l/R and R-curve characteristics on the fracture behaviour of hardmetals is here recommended for a correct determination of the fracture toughness of cemented carbides by means of the SEVNB method. This consideration is particularly relevant for the case of relatively high-toughness grades, where a sharp pre-crack may not be easily induced.

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