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Mechanical deformation of WC–Co composite micropillars under uniaxial compression

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ABSTRACT

In this work, WC–Co micropillars machined by focused ion beam have been tested under uniaxial compression to investigate the stress–strain behavior and associated deformation mechanisms. The results indicate that yielding phenomena is evidenced by multiple strain bursts. Experimental data is found to fall within the bounds defined by the mechanical responses expected for an unconstrained Co-binder like model alloy and a bulk-like constrained binder region in WC–Co composites; capturing then local phase assemblage and crystal orientation effects.

1. Introduction

Understanding the deformation of ductile metallic ligaments constrained by a hard ceramic phase is a key feature for determining effective toughening in metal reinforced ceramic-base composites [1–3]. WC–Co cemented carbides, also referred to as hardmetals, are one of the most extensively employed "tailor-made" composites and a clear example of this type of material. It consists of two interpenetrating-phase networks (i.e., soft/ductile metallic binder and hard/brittle ceramic particles), where toughening through constrained deformation of the ductile phase is highly effective. As a consequence, they exhibit outstanding toughness levels as related to the energy required to plastically deform the metallic bridging ligaments that develop behind the tip of preexisting or service-induced cracks [4–9]. During the last decades, extensive effort has been devoted to predict the mechanical response of the constrained ductile bridges and its contribution to the fracture toughness of the composite. As a result, several models have been presented [1,2,5,7,10]. However, these models are based on the macromechanical response of the composites investigated, and more important, are limited due to the scarce information on the interactive deformation of the constitutive phases at the microstructural length scale.

High toughness levels of cemented carbides have also been invoked as the result of the effective interaction between the intrinsic residual stress state and the external applied stress [11–15]. In this regard, it is known that very large thermal residual stresses (TRS) develop in these materials as they are cooled from sintering temperature, due to the large difference between the thermal expansion coefficients of both constituent phases. The thermal expansion coefficient of Co is more than twice that of WC, and therefore high tensile TRS can be expected in the metallic phase, while the overall stress state of the WC particles is compressive [11]. However, wide stress ranges have been evidenced in the local stress states of both phases, including tensile stresses at some WC angularities [11]. Under the application of external loads, TRS interact with the applied stress field and strongly influence the plastic deformation of the system. During uniaxial compressive loading, main plastic deformation is accommodated by the binder phase; although small plastic strains have been also evidenced in WC particles [12]. In the axial direction, mean TRS in the binder phase oppose the compressive applied stress, but in the transverse direction the tensile stress state that results from the Poisson effect adds to the TRS resulting in binder flow in this direction [12, 14]. Nevertheless, addressing the exact stress-state produced under the application of compression loads is extremely complicated because the macroscopic plasticity is the result of the aggregation of multiple micro-scale yield events within particular phases. Furthermore, during loading the strain and stress states of the system are in constant evolution due to the continuous redistribution and cancelation of internal plastic strains between phases [12].

Recent advances in micro- and nano-fabrication and testing systems have enabled the assessment of deformation behavior of bulk materials on a microscopic scale. The use of focused ion beam (FIB) technique combined with nanoindentation has provided the means for machining

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and uniaxial compression testing of micropillar samples [16]. Within this context, while significant effort has been dedicated to study the mechanical behavior of single-crystals or boundary-containing metallic systems [17], less attention has been paid to the case of composite materials (e.g., Refs. [18,19]), and particularly those combining soft and hard phases (e.g., Refs. [20,21]). To the best knowledge of the authors, experimental micromechanics studies on WC–Co composites reduce to two recent reports [22,23]. In one of them, micropillar compression testing of hardmetals was conducted by Csanádi et al., but their investigation was limited to evaluation of the deformation characteristics of the hard carbide phase [22]. On the other hand, Trueba and coworkers documented and analyzed fracture events in WC–Co cemented carbides by microbeam testing and finite element modeling [23]. It is the aim of this study to bring insights on the mechanical deformation and failure behavior of ceramic–metal composite materials through the compression of micropillars consisting of Co-binder ligaments constrained by their surrounding WC carbides. In doing so, special attention is paid to document and analyze microstructural effects regarding yield strength and constraining degree.

2. Experimental procedure

The micropillars were carved into the surface of a coarse-grained WC–15%$_{wt.}$Co composite by means of a Zeiss Neon 40 FIB milling system operated at 5 kV. Milling process was carried out in two stages in order to minimize damage by impinging ions. Initially, a ring with outer and inner diameters of 15 μm and 4 μm was carved using an ion beam current of 4 nA. Afterwards, micropillars with final diameters varying from 2.5 to 3 μm, aspect ratios ranging from 2 to 2.5 and taper angles between 2° and 3°, were shaped using a 500 pA current. Two examples of micropillars before compression tests are shown in insets within Fig. 1. The micropillars were uniaxially compressed using a Nanoindenter XP (MTS) fitted with a 5 μm diameter flat diamond-punch at a constant displacement rate of 10 nm/s (initial strain rates around 0.0015 s^{-1}). Load–displacement data was continuously recorded in the same way as being practiced in nanoindentation measurements. Nominal stresses and strains were directly determined from the load–displacements curves, using the diameter at one quarter of the way down the pillar (as most deformation occurred in this region) and its effective gauge length, respectively. Four micropillars were indented at different depths corresponding to maximum axial strains (ε_f) of 2.9, 3.8, 4.4 and 5.5%. Irreversible deformation and failure mechanisms have been directly examined by means of Field Emission Scanning Electron Microscopy (FESEM), as well as by serial sectioning and imaging using the FIB/FESEM system.

3. Results and discussion

FESEM micrographs of two micropillars compressed up to maximum strains of 2.9 and 4.4% are shown in Fig. 1, before (insets) and after compression tests. Two different shearing mechanisms are identified in the FESEM micrographs. The first one (very clear in Fig. 1b) takes place at the interface between the WC particle and the metallic binder, at angles comprised between 30° and 45° with respect to the compression axis. Detailed analysis permits to discern that shearing does not occur exactly at the interface, but rather proceeds within the binder very close to the phase boundary and parallel to it. This finding is consistent with the fact that binder regions adjacent to carbide/binder interfaces are preferred crack growth locations, due to coincidence of high plastic strains and maximum triaxiality conditions [6]. The second deformation mechanism develops at the grain boundaries between contiguous WC crystals. Although it may be speculated that observation of different shearing/cracking mechanisms should be dependent upon specific crystal orientation and local phase arrangement (i.e., effective constraining degree) within the pillar, it is clear that interfaces, between either binder and carbide or carbides themselves, are favorable points for driving irreversible deformation and failure events under macroscopic compressive stresses.

To better understand the deformation/failure mechanisms under uniaxial compression tests, the micropillar compressed up to a strain of 4.4% was sequentially cross sectioned and visualized using the FIB/FESEM system. Thus, micrographs corresponding to the interior of the deformed pillar are shown in Fig. 2 where the most prominent events are marked with white arrows. In all three micrographs a glide system within the binder adjacent to the interface with the WC particles can be appreciated (at an angle of about 42° with respect the compression axis). Furthermore, in the central image (Fig. 2b), a microcrack running parallel to the carbide/binder interface (but still within the binder phase) is also identified. This microcrack probably stems from the propagation of the carbide–carbide interface microcrack that is at the same position in Fig. 2a. Binder regions close to carbide corners combine large concentrations of strains and/or stress triaxiality; thus, they are favorable zones for early flow and/or crack propagation [6]. On the other hand, all carbide–carbide interfaces are affected by compression loading, pointing out that they are weak links in these ceramic–metal composites [5]. Evidence of extensive plastic deformation within the binder is observed in Fig. 2c. It may come from less effective constraining or plastic flow associated with local interaction between TRS and applied stress. In this regard, it should be pointed out that the already complex point-to-point residual stress state in both phases may be relaxed or enhanced by local tensile stresses related to the Poisson effect under the nominally compressive applied one [12,14].

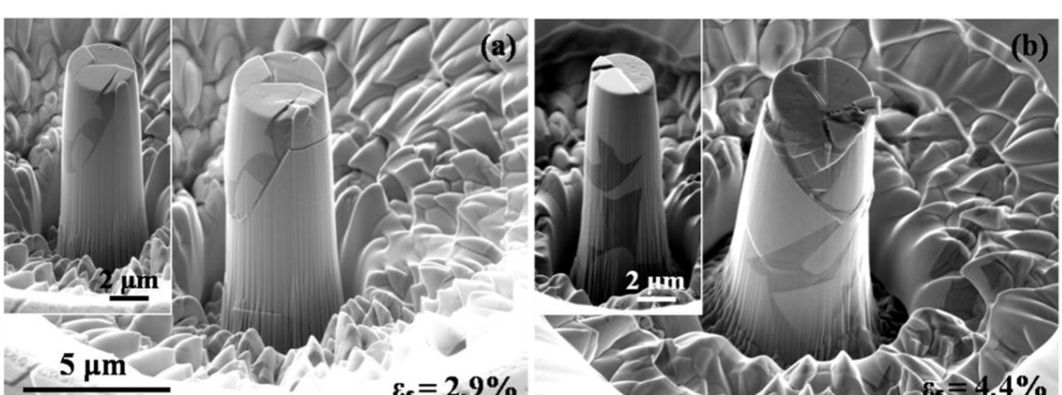

Fig. 1. FESEM micrographs of compressed micropillars up to maximum axial deformations of (a) 2.9% and (b) 4.4%. The insets show the appearance of the micropillars before compression. Binder–carbide (within the binder phase) and carbide–carbide interfaces are the weakest points for failure.