Synthesis and characterizations of ball-milled nanocrystalline WC and nanocomposite WC–Co powders and subsequent consolidations

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Abstract

The room-temperature ball-milling technique has been successfully employed for fabrication of nanocrystalline powders of equiatomic WC by high-energy ball milling of elemental W and C powders. The progress of the solid state reaction has been monitored by means of X-ray diffraction, scanning electron microscopy and transmission electron microscopy (TEM). A complete single phase of hcp-WC was obtained after 295 ks of milling. Increasing the milling time to 432 ks leads to dramatic grain refinement and the powder of this end-product consists of nanocrystalline grains of less than 5 nm in diameter. Part of the powders of the end-product were further ball-milled, 259 ks together with different concentrations of metallic Co to obtain spherical fine nanocomposite WC-coated Co powders with an average particle size of less than 5 μm in diameter. Both WC free Co and composite WC–Co powders were then consolidated into fully dense (>99.5%) compacts using cold and hot pressing techniques. The as-consolidated samples still maintained their nanocrystalline characteristics with an average grain size of less than 100 nm. Some of physical and mechanical properties of the consolidated samples are reported. © 2000 Elsevier Science B.V. All rights reserved.

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1. Introduction

Nanocrystalline materials [1] that are defined as materials with grain sizes less than 100 nm have received much attention as advanced engineering materials with improved physical and mechanical properties [2,3]. Mechanical alloying (MA) [4] which is a well known process for preparing several amorphous [5–10], metal nitrides [11,12], metal carbides [13–16] alloys, and nanocomposite materials [17], has been considered as a powerful technique for synthesizing numerous nanocrystalline materials. Recently, MA process has become a popular method to fabricate nanocrystalline materials due to its simplicity and relatively inexpensive equipment [18,19]. Their unusual unique properties make transition-metal carbides very interesting and important for several industrial applications. They have excellent high temperature strength and good corrosion resistance, being chemically and thermally very stable even at the high temperatures. They are extremely hard materials and possess high values of Young’s modulus. In addition, they are typically metallic in their electrical and optical properties. Among the hard alloys and refractory carbides, WC finds wide range of industrial applications, being used extensively in commercial applications such as tips for cutting and drilling tools, wear resistant parts in wire drawing, extrusion and pressing dies and wear-resistant surfaces in many types of machines.

Since the hexagonal phase of WC is usually obtained by direct union of elemental W and graphite at temperatures ranging from 1673 to 1873 K [20,21], their preparation costs are very high. The present study proposes an attractive technique for preparation of high quality powders of nanocrystalline WC at room temperature, by ball-milling a mixture of W and graphite powders. The as-synthesized nanophasic WC has been milled with selected...
concentration of metallic Co (4, 8 and 14 wt.%) resulting cemented W–Co nanocomposite materials. The end-product of the milled powders for both WC and WC–Co are then consolidated into fully dense compact, so that the density and many mechanical properties could be determined. One aim of the present work is to offer an inexpensive technique for fabrication such technological important materials, using a simple and powerful method that can be easily scaled up to produce larger quantities of several kilograms, using larger types of mills, e.g. low energy ball or rod mills [22,23].

2. Experimental details

Elemental powders of high purity (99.9%) W (75 mesh) and C (350 mesh) were used as starting reactant materials. The powders were mixed to give the desired average composition of equiatomic WC and sealed in a stainless steel (SUS 316) vial (80 ml in volume) together with 10 stainless steel (SUS 316) balls (12 mm in diameter) in a glove box under Ar gas atmosphere (O2 and H2O are less than 10 ppm). The ball-to-powder weight ratio was selected to be 10:1. The milling procedure was carried out at room temperature, using a high-energy ball mill (FRITACH P6). The milling process was interrupted after selected milling times, and a small amount of the ball-milled powder was taken from the vial in the glove box for several analyses. The final product of the milled powders (hcp-WC) was then mixed with selected concentrations (4, 8 and 14 wt.%) of metallic hcp-Co (150 mesh, 99.9%) and ball-milled for 86 ks using the same milling apparatus that was used for preparing the WC powders. In order to avoid increasing contamination of iron which usually comes from the using steel milling tools, agate milling tools have been used to prepare the powders of WC–Co composites. X-ray diffraction (XRD) with CuKα radiation, transmission electron microscopy (TEM) operated at 200 kV, have been used to monitor the structural changes of the powders after several ball milling times. However, some samples have been characterized by high-resolution transmission electron microscopy (HRTEM) operated at 200 kV. The samples of the TEM and/or HRTEM have been prepared by mixing the powders with small amounts of alcohol (2 ml in volume) and stirring for 30 s. Two or three drops of the suspension were dropped on a Cu-microgrid and then well dried for about 1.8 ks before mounting the microgrid onto the TEM sample holder. The morphological (shape and size) changes of the powder after selected milling times were determined by scanning electron microscopy (SEM) operated at 15 to 35 kV. In order to determine the concentrations of W, Co and the contamination of Fe, the WC and WC–Co powders of the final product have been analyzed by the induction coupled plasma emission method. On the other hand, the concentration of C, and the degree of oxygen contamination that is introduced into the powders during the milling procedure and/or during the powder handling outside the glove box, have been determined by the helium carrier fusion-thermal conductivity method.

The powders of the end-products of WC and WC–Co were consolidated into compact samples, using two steps. In the first step, the powders were cold-compacted at room temperature under purified Ar atmosphere, using a tungsten carbide die of 13.0 mm in diameter at a pressure of 1.0 GPa for 70.2 ks. The obtained green-compact which has a relative density of about 60% of the theoretical density was then transferred to another tungsten carbide die of 15 mm and then hot-pressed in vacuum under a pressure of 1.5 GPa at 1973 K for 43 ks without the addition of any binding materials.

The density of the consolidated samples was determined by Archimedes’ principle, using water immersion. The hardness of the as-consolidated samples was determined using a Vickers hardness indenter with a load of 30 kg. The hardness values reported below (Section 3) are averaged from ten indentations. Moreover, the bulk samples were structurally analyzed using XRD and TEM techniques. In addition, the samples were examined metallographically by a light optical microscope and SEM/EDX techniques. Some mechanical properties of the consolidated samples have been determined by non-destructive tests.

In order to know whether the level of contamination has been increased or not after the consolidation procedure, chemical analyses were preformed for the hot-pressed samples

Table 1 shows the chemical analyses of the as-milled and as-consolidated WC samples. Obviously, the as-hot pressed sample gains 0.38 at.% oxygen during the consolidation procedure, although, the hot pressing was achieved under high vacuum, as high as 6.0×10⁻⁶ Torr. This may be attributed to some leaks in the vacuum system of the hot press. It is worth noting that when metallic Co powders are added to the final product of WC, the oxygen contamination content in the bulk samples is increased to the level.

| Sample                          | W content (at.%) | C content (at.%) | Co content (at.%) | Fe content (at.%) | O content (at.%)
|---------------------------------|-----------------|-----------------|------------------|------------------|----------------
| As-fabricated WC powders        | 49.40           | 48.70           | -                | 1.45             | 0.45           |
| As-consolidated WC              | 49.2            | 48.52           | -                | 1.45             | 0.83           |
of 1.44 at.%. This is attributed to the action of the fresh surfaces of metallic Co to gutter the oxygen. Nevertheless, the total contamination content of the bulk sample is 2.28–2.89 at.%, being acceptable for most industrial applications.

3. Results

3.1. Formation of nanocrystalline WC powders

3.1.1. Structural changes with milling time

The XRD patterns of ball-milled W–C binary system at the starting (0 ks), early (22–130 ks) and intermediate (173 ks) stages of milling are displayed together in Fig. 1. Whereas Fig. 2a and b show the XRD patterns of the final products at the final stage of milling (256–432 ks). The starting reactant materials (Fig. 1a) consist of coarse polycrystalline grains of W and carbon, indicated by the sharp Bragg diffraction peaks (Fig. 1a). During the first few kiloseconds of milling, exemplified by 22 ks (Fig. 2b) a new phase appeared (labeled by closed circles). The Bragg diffraction lines of this phase become more pronounced upon further milling (130–173 ks), suggesting an increasing of the volume fraction of this new phase against Bragg reflections that come from the pure elemental W and the reactant materials of the starting powders. The X-ray analysis that is performed to identify this phase formed during these stages of milling suggested the formation of hcp-WC powders. After 256 ks of milling (Fig. 2a), all Bragg reflections that come from the pure elemental W and C crystals vanished and the XRD patterns reveal a single phase of hcp-WC with an average grain size of about 70–90 nm in diameter, as presented in Fig. 3 which shows the bright field image (BFI) of the milled powder at this stage of milling. Increasing the milling time (432 ks) leads to further grain refining of the formed product of WC to obtain nanocrystalline hcp-WC, as presented in Fig. 2b. The BFI and the HRTEM of this final product shows the formation of nearly equiaxed nanocrystalline grains of WC with grain sizes ranging from 5 to 7 nm diameter, as presented in Fig. 4a and b, respectively. Neither unprocessed W nor C crystals could be detected, indicating the completion of the solid state reaction and the formation of a single nanophase of hcp-WC.
sharp edges of an average size of 97 nm in diameter. The SADP of this sample reveals a ring-like diffraction pattern without the existence of sharp spots which also indicates the formation of ultrafine grains (Fig. 5b). Neither unprocessed reactant materials nor other carbide phase(s) could be detected, and the SADP reveals an hcp-structure corresponding to the WC phase, as shown in Fig. 5b.

3.1.2. Morphology change with ball-milling time

There can be no more fundamental characteristics of a powder than the size and the shape of the individual particles. Thus, the SEM technique was used to follow the change in size and shape of the ball-milled WC powders after several milling times. The SEM observations for the powders that were taken after 22 ks of milling (early stage) are bulky and have flake-like morphology with a wide size distribution, ranging from 30 μm to nearly 90 μm in diameter, as shown in Fig. 6a. In contrast to the powders of the early stage of milling, the powders at the second stage (173 ks) tend to have regular shape (nearly spherical) with a narrow size distribution, ranging from 7 to 25 μm in diameter (Fig. 6b). Towards the end of the ball-milling time (432 ks, final stage) the powders reveal ultrafine characteristics with about 0.5 μm (or less) in diameter of uniform equiaxed spheres with smooth surfaces, as presented in Fig. 6c.

Fig. 7 shows the cross-sectional view of as-consolidated WC powders that were obtained after 432 ks of milling. This sintering step that was performed at 1973 K leads to the formation of uniform grains with an average diameter of less than 10 μm in diameter. The irregularities, which can be seen at the grain edges, are caused by the excessive etching during the metallographic examinations.

Fig. 8 presents the correlation between the consolidation temperature, the relative density (closed symbols) and grain size of the consolidated samples (open symbols). The relative density of the powders that were consolidated at room temperature have a density of about 58% (green compact), with an average grain size (7 nm in diameter) which does not differ markedly from the milled samples (5–7 nm in diameter). Increasing the consolidation temperature to 1073 K leads to a marked increase the relative density to about 78%. Accordingly, the grains have been grown in size to nearly 65 nm in diameter. The bulk sample that was consolidated at 1273 K has higher density of about 84%, with an average grain size of 70 nm. This relative density has increased to be more than 90% upon consolidation at 1473 K and 1673 so that the grain sizes have increased to values ranging from 85 to 90 nm. Nearly fully-dense (above 99.5%) samples could be obtained at consolidation temperatures ranging between 1873 and 1973 K. Unfortunately, consolidation of the powders at such high temperature leads to a marked grain growth so that the bulk compact sample consists of larger grains of about 94–97 nm in diameter.

A well-known empirical dependence of strength and
hardness on porosity and grain size for ceramics materials is given by [15, 25–27]

\[ H = Kd^{-a} e^{-bP} \]

where \( H \) is the hardness value, \( d \) grain size, \( P \) specimen porosity, and \( K, a, \) and \( b \) empirical constants. The constant \( a \) can be approximated to be 0.5 [15, 29].

Fig. 9 shows the correlation between the \( P \) of the consolidated samples that were hot-pressed at various temperatures, and the combined effect of \( H \) and \( d \). It can be noted that the hardness increases with increasing density (low porosity values) and grain size. A linear relation is obtained with a slope, \( b = 5.351 \), similar to that found for a variety of materials (3.7 [25–27] 6.6–7 [28, 29] 5.575 [15]).

3.2. Formation of nanocomposite WC–Co powders

In order to improve the toughness of the mechanically solid state synthesized WC brittle materials, elemental \( hcp \)-Co powders with selected weight percentages (4, 8 and 14 wt.%) have been mixed ball-milled with the fabricated nanocrystalline powders of WC (see Section 2).

Fig. 10 displays the SEM micrograph of mechanically ball mixed WC–14 wt.% Co after 86 ks of ball-milling. At this stage of milling, the ductile powders of Co are deformed and somewhat agglomerated so that the hard WC particles are embedded into the Co powders to form particles of WC-coated by Co particles that have rather irregular surface relieves. Those small particles (under 1 \( \mu m \) in diameter) are Co-free WC particles (Fig. 10).

Fig. 11 presents the HRTEM micrograph of the composite WC–14 wt.% Co after 86 ks of ball-milling. Obviously, the nanocrystalline WC particles (fringe imaged dark gray equiaxed grains) are embedded and homogeneously distributed into the matrix of \( hcp \)-Co (light gray matrix) to form nanocomposite WC–Co materials with the absence of any reactive products.

The powders of nanocomposite WC–14 wt.% Co materials were then hot-pressed at 1973 K to form bulk nanocomposite materials. Fig. 12 shows the back scattering SEM micrograph for the cross-sectional view of the consolidated powders (Fig. 12a) and the X-ray maps of elemental W (Fig. 12b) and Co (Fig. 12c). From the cross-section of the consolidated sample it is possible to establish that the fine angular WC grains (Fig. 12b) are embedded in the binding continuous matrix material of Co (Fig. 12c). Obviously, the intensity of the dots corresponding to particular W and Co particles segregated in proportion to their concentration at a given location. One can see a homogeneous and uniform WC distribution into the host matrix of metallic Co.

The BFI and the corresponding SADP of the as-consolidated WC–14 wt.% Co materials that was hot-pressed at 1973 K, are shown in Fig. 13a and b, respectively. The micrograph (Fig. 13a) shows again a homogenous distribution of the nanograins WC (~95 nm white angular grains into the continuous binding matrix of Co, dark gray matrix). A high density of the sintered sample can be remarked. The SADP (Fig. 13b) revealed that the consolidated powders contain \( hcp \)-WC and \( hcp \)-Co with no other phases.

The bulk densities of the nanocomposite WC–Co materials were determined by Archimedes’ principle using water immersion and are plotted in Fig. 14 as a function of Co content. Two mathematical expressions have been used to predict the effect of Co content on the density of these composite materials and to compare them with the measured values. These rules of mixture equation [30] show that the measured densities of the fabricated samples for nanocomposite WC–Co fall between an upper bound presented by

\[ \rho_{WC-Co} = \rho_{WC} V_{WC} + \rho_{Co} V_{Co} \]
Fig. 7. Cross-sectional view of the as-consolidated W_{60}C_{40} powders milled for 432 ks.

Fig. 8. Correlation between the relative density and the average grain size of as-consolidated W_{60}C_{40} (432 ks of the ball milling time) at several temperatures of hot-pressing.

and a lower bound, or limit,

\[ \rho_{\text{WC}} \rho_{\text{Co}} \rho_{\text{WC}}^V + \rho_{\text{Co}} \rho_{\text{WC}}^V \]

Here, \( \rho \) and \( V \) denote the density and the volume fraction, respectively.

A polarized light optical micrograph of Vickers hardness indentation of the consolidated WC–14 wt.% Co samples displayed in Fig. 15. The hardness of this compact sample was measured to be 11 GPa, taking the average of at least ten indentations. The cracks displayed in the micrograph that developed in the product during indentation and extended 100 to 150 \( \mu \text{m} \) away from the it, were used as indicator for estimating the fracture toughness, \( K_c \), via a model suggested by Anstis et al. [31]. This value is found to be about 14 MPa.m^{1/2}, well above that of commercial WC/Co composites (12 MPa.m^{1/2}) [32]. This indicates the possibility of producing higher fracture toughness WC–Co nanocomposite materials.

The dependence of the hardness and fracture toughness of the composite WC–Co on additives of elemental Co (wt.%) is displayed in Fig. 16. The starting material of pure nanocrystalline WC powders that were obtained after 432 ks of milling elemental W and C has an exceptional
Fig. 9. Dependence of the Vickers hardness on porosity and grain size of bulk W₆₀C₄₀ (432 ks of ball milling).

Fig. 10. SEM micrograph of mechanically mixed W₆₀C₄₀ with metallic Co (14% wt.%) powders after the ball milling time.

Fig. 11. HRTEM micrograph for mechanically mixed W₆₀C₄₀ with metallic Co (14% wt.%) powders after ball milling.

hardness as high as 21 GPa. However, it shows a poor toughness value of less than 1 MPa.m¹/². Addition of ductile Co improves its toughness so that the latter is monotonically increased with increasing volume fraction of metallic Co. Contrary to this, the hardness is significantly decreased with increasing the ductile phase to reach a value of 11 GPa (14 wt.% Co). It is worth noting that this value is higher than that of traditionally fabricated WC–14 wt.% Co (10.2 GPa) [32].

Poisson’s ratio and the elastic modulii of the consolidated powders have been estimated from the measured sample densities and the constant parameters of the nondestructive testing apparatus. These are plotted as a function of the Co content in Figs. 17 and 18, respectively. Increasing the content of the ductile versus the brittle phase of WC leads to a monotonic increase in the value of Poisson’s ratio to be nearly 0.235 for nanocomposite WC–14 wt.% Co. This value is larger than that of pure nanocrystalline WC (0.005), suggesting an improvement in the ductile characteristics of the fabricated materials.

The elastic modulii presented by Young’s modulus and the shear modulus of the consolidated WC–Co powders are shown as a function of the Co content in Fig. 18. These values decrease significantly with increasing weight fraction of the ductile Co to reach values of 540 GPa and 220 GPa, respectively. This is in good agreement with the values (525 GPa and 210 GPa) that are obtained by the traditional way of fabrication [32].

4. Discussion

Tungsten carbide is one of the hardest, stiffest and most refractory materials known and these open for WC a wide range of applications as refractory material. In spite of the commercial methods for preparing WC refractory materials, it has been demonstrated by El-Eskandarany et al. [33] that there exists the possibility of forming nanocrystalline WC by high-energy ball milling from a mixture of WO₃, Mg and C powders at room temperature, using a so-called mechanical solid state reduction process [34]. The end-product is then leached to remove the residual MgO and to obtain nanoscale grains of WC powders that were consolidated into a nanocrystalline compact sample [35]. When the MgO remained in the milled powders, a new nanocomposite materials of WC–18 at.% MgO with advanced mechanical and physical properties has been produced [36]. In the present work, we have proposed another powerful and simple process for the formation of nanocrystalline WC powders at room temperature. The
4.1. Fabrication of nanocrystalline WC by milling of elemental W and C powders

Based on the results of the present study, the fabrication of WC powders via the ball-milling route involves three milling stages. During the first stage of ball-milling (0 to 130 ks) a small mole fraction of hcp-WC has been formed (see Fig. 1b,c) as a result of the mechanically induced solid state reaction that took place between the reactant materials of metallic W and C powders. At this stage of milling, the powders are heterogeneous in both shape and size (see Fig. 6a). Increasing the milling time to 173 ks (second stage of milling) leads to an increase of the impact and shear forces that are generated by the milling tools (balls) and this enhances the solid state reaction between the diffusion couples of unprocessed W and C powders. At this stage of milling, all the C atoms are hardly seen, suggesting a solid state solubility of the small C atoms in the W lattice to form a solid solution of bcc-WC (Fig. 1d) coexisting with another product phase of hcp-WC already formed in the previous stage of milling. The lattice parameter of the bcc-W was calculated to be 0.31712 nm, being larger than that one for pure metallic bcc-W (0.31648 nm [37]) and indicating an expansion of the W lattice and the formation of solid solution. Since this formed solid solution is a metastable phase it cannot withstand the shear and impact forces that are generated by the balls so that it transforms into the most stable phase of hcp-WC during the final stage of milling, 256 ks (Fig. 2a). In spite of these results, Mi et al. [38] have attributed the absence of the C Bragg peaks in their milled powders of W and C to the low scattering factor of C or to amorphization. However, they obtained hcp-WC with a residual metallic phase of W upon post annealing of pre-milled W and C powders, using the HIPing consolidation technique.

At the beginning of the final stage of milling, the obtained powders of pure WC have nearly spherical like morphology with fine particle size (Fig. 6b). This end-product consists of nanocrystalline spherical lenses of less than 100 nm in diameter (Fig. 3). Increasing the milling time to 432 ks does not lead to any phase transformations (Fig. 2b) and the end-product of the powders that have excellent morphological characteristics (Fig. 6c) consists of extremely fine WC grains [less than 7 nm in diameter, (Fig. 4)].

The formation of this WC nanocrystalline material can be attributed to the plastic deformation that is usually produced in crystal lattices during the high-energy ball milling process and this occurs by slip and twinning in the lattice of the milled powders. Accordingly, this deformation involves localization in shear bands containing a high dense network of dislocation. Due to the successive accumulations of the dislocation density, the crystal are disintegrated into subgrains that are initially separated by low angle grain boundaries. The formation of these subgrains is attributed to the decrease of the atomic level
strain. Further ball milling time leads to further deformation occurring in the shear bands that are located in the unstrained parts of the powders. These leads to subgrain size reduction so that the orientation of the final grains becomes random with respect to the crystallographic orientations in the numerous grains and hence, the direction of slip varies from one grain to another [39].

Reduction in grain size is a very important factor for consolidation because it increases the sinterability of the powders and improves the mechanical and physical properties of the sintered materials (Figs. 8 and 9). In the present work, a hot-pressing technique has been used to consolidate the nanocrystalline WC powders into a fully-dense bulk compact. This consolidation step is necessary for most industrial applications. One can say that the grains of the milled WC powders have grown by a factor of 14 upon this consolidation step. This grain growth is still acceptable since the sintered sample of WC is nearly fully-dense (Fig. 8) and still maintains its advanced nanocrystalline characteristics (about 97 nm in diameter; Figs. 2c and 5) with extreme hardness value (Fig. 16).

4.2. Fabrication of nanocomposite cemented WC–Co by milling of WC and Co powders

However, the previously consolidated nanocrystalline WC materials have high elastic modulii (Young’s and
fracture, the mechanically induced solid state fabricated WC nanocrystalline of the present work have been mixed with selected concentrations of metallic Co powders. The combination of WC and metallic Co as a binding matrix material is a well-adjusted system not only with regard to its properties, but also to its sintering behavior. In fact, Co is a desirable material because it can wet the surface of WC upon ball-milling (Fig. 10) while it maintains negligible solubility in the carbide phase.

When the as-mixed WC–Co powders were hot-pressed (liquid phase sintering) at a temperature (1973 K) above the melting point of Co (1768 K), the liquid Co which primarily wetted the surfaces of WC during the milling (Fig. 10) could react with the surface of the WC powders. During the cooling of the hot-pressing procedure, the liquid phase of Co is therefore cooled and freezes (crystallized) so that the WC is trapped into the host matrix of the binding frozen Co and welded tightly together to form the composite structure. This leads to the formation of pore-free structure composite WC–Co materials (Figs. 12–14).

It is worth noting that the sintered WC–Co materials consists of nanocrystalline WC (reinforcement material) embedded in a nano-continuous matrix of metallic Co (Fig. 13), forming nanocomposite WC–Co. Increasing the amount of metallic Co in the milled WC powders leads to a marked decrease in the elastic moduli (Fig. 18). However, it improves the other mechanical properties such as Poisson’s ratio (Fig. 17). In addition, the fabricated WC–Co nanocrystalline materials have high enough fracture surface and hardness values (Figs. 15 and 16), far being suitable as tools and dies.

5. Conclusions

In the present study, elemental W and C powders have been milled together at room temperature using high-energy ball milling. A single phase of hcp-WC powders was obtained after ball-milling for 259 ks. The powders of the end-product (432 ks of milling) possess excellent morphological properties (spherical-like morphology with diameters less than 0.5 μm). The powders of this stage of milling that consist of nanocrystalline grains (less than 5 nm in diameter), were then consolidated into compacts, using cold and hot pressing techniques. However, a marked grain growth of the consolidated materials has taken place during the hot-pressing procedure, the compacted samples still maintaining their unique nanocrystalline characteristics (97 nm in diameter). The as-consolidated samples are fully dense (15.36 g/cm³), having more than 99.5% of the theoretical density of WC. The results of the Vickers hardness tests have indicated a significant strengthening effect due to grain size refinement and a monotonic hardness increase with decreasing residual porosity. Contrary to the obtained extremely high values of
hardness (21 GPa) and elastic modulii (indexed by Young's modulus, 800 GPa and shear modulus, 400 GPa), the fabricated WC shows poor values of fracture toughness (about 1 MPa.m\(^{1/2}\)) and Poisson’s ratio (0.001). In order to improve the toughness characteristics, metallic Co powders were mixed with the end-product of nanocrystalline WC powders and mechanically-ball milled for 259 ks. During this solid-state mixing, the powders of Co wetted the surfaces of the WC particles to form WC-powders that are coated by Co. The as-mixed composites WC–Co were then hot-pressed to form dense crenate nanocomposite (nearly 95 nm in diameter) materials. Increasing the concentration (14 wt.%) of the binding continuous matrix of Co leads to increase the fracture toughness and Poisson’s ratio of the WC (14 MPa.m\(^{1/2}\) and 0.235, respectively). Our work propose room-temperature ball milling for fabrication of large amounts of powders of metal carbides and composite materials. These can be used as starting materials for nanocrystalline and nanocomposite materials and after subsequent consolidation into fully dense nanophase bulk materials.

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