Chemically vapour deposited diamond coatings on cemented tungsten carbides: Substrate pretreatments, adhesion and cutting performance

Riccardo Polini

Dipartimento di Scienze e Tecnologie Chimiche, Università degli Studi di Roma Tor Vergata, Via della Ricerca Scientifica, 1 - 00133 Roma, Italy

Available online 18 January 2006

Abstract

Chemical vapour deposition (CVD) of diamond films onto Co-cemented tungsten carbide (WC-Co) tools and wear parts presents several problems due to interfacial graphitization induced by the binder phase and thermal expansion mismatch of diamond and WC-Co. Methods used to improve diamond film adhesion include substrate-modification processes that create a three-dimensional compositionally graded interface. This paper reviews substrate pretreatments and adhesion issues of chemically vapour deposited diamond films on WC-Co. The combined effect of pretreatments and substrate microstructure on the adhesive toughness and wear rate of CVD diamond in dry machining of highly abrasive materials was analyzed. The role of diamond film surface morphology on chip evacuation in dry milling of ceramics was also investigated by comparing feed forces of coated and uncoated mills. The overall tribological performance of diamond coated mills depended on coating microstructure and smoothness. The use of smoother films did allow to reduce cutting forces by facilitating chip evacuation.

© 2005 Elsevier B.V. All rights reserved.

Keywords: Chemical vapour deposition (CVD); Diamond; Tungsten carbide; Cutting tools

1. Introduction

Since the pioneering report of diamond deposition onto non-diamond substrates using the gas phase by Derjaguin and Fedoseev [1] in 1975 and the confirmation provided by Matsumoto et al. [2] in 1982 using hot filament chemical vapor deposition (HFCVD) technique, diamond film technology has attracted many research groups around the world to study new synthesis methods and applications as well as fundamental issues concerning nucleation and growth mechanisms [3–6].

Diamond film deposition has thus made it possible to use many diamond properties [7] in new and innovative applications. Diamond is widely used for tooling applications in the form of crystals within a matrix. However, the surprisingly simple procedures for producing low-pressure diamond films now means thin films can be obtained on wear components having complex geometries. The ultimate goal of using CVD diamond coatings is to extend the technical performance of components, such as tools, beyond their conventional wear life.

Coating wear resistance is not the only parameter limiting the performance of CVD diamond coated tools. When properly synthesized, CVD diamond films are highly resistant to chemical and abrasive wear. Moreover, diamond-coated articles are of little use if these deposits do not remain attached to the underlying substrate during use. Consequently, in the case of CVD diamond coated articles it is mandatory to optimize both adhesion and wear resistance.

The most suitable substrate material for producing diamond coated tools is Co-cemented tungsten carbide (WC-Co). In fact, although diamond is in direct contact with the material to be machined, given the limited thickness of the deposit, mechanical stresses are transmitted to the substrate. Therefore the direct deposition of CVD diamond onto WC-Co inserts would result in a cutting tool which is extremely resistant to abrasion while retaining high fracture toughness in the underlying WC-Co substrate. However, premature adhesion failure of CVD diamond coatings is the main issue preventing useful application of diamond films as wear resistant coatings for carbide tools.

Co-cemented carbides contain cobalt as a binder, which provides additional toughness to the tool, but it is hostile to diamond adhesion [8–10]. According to the phase diagram of the Co-C system, at typical diamond CVD temperatures (700–1000 °C) carbon is soluble in Co up to 0.2–0.3 wt.% C [11]. During the initial stages of the CVD process, the WC-Co is exposed to a hydrocarbon radical rich atmosphere and carbon species can rapidly diffuse into the bulk of the binder phase.
Fig. 1. SEM micrograph of the substrate side of a delaminated diamond film. Failure occurred by crack extension below the outermost layer of binder-free WC grains.
More recently, the author of this paper studied the use of CrN as an interlayer for diamond deposition on WC-Co. Substrates used were 10 × 10 × 4 mm WC-5.8 wt.% Co as-ground samples (5 μm WC grain size, supplied by Fabbrica Italiana Leghe Metalliche Sinterizzate S.p.A., Anzola d’Ossola, VB, Italy). A PVD-arc plant (Microcoat S.p.A., Italy) was used to deposit CrN thin films (7 Å). PVD coated substrates were prepared before diamond deposition in order to enhance nucleation. Samples were ultrasonically seeded for 15 min in 1/4 Å diamond suspension (Struers, DP-Suspension, HQ), then cleaned with ethanol in an ultrasonic bath. Diamond films were deposited for 9 h in a stainless steel hot-filament chemical vapor deposition (HFCVD) chamber described elsewhere [28]. The gas phase was composed of a mixture of methane and hydrogen with a CH₄/H₂ volume ratio fixed at 1.0% and was activated using a tantalum filament (2170 °C). The total pressure of the gas mixture in the reactor was 4.8 kPa and flow rate was 300 sccm (standard cm³ min⁻¹). Substrate temperature was 650 °C and diamond film thickness was 6 μm.

Grazing incidence (1°) X-Ray Diffraction (XRD, Philips X’Pert Pro, Cu Kα radiation) was used to characterize CrN interlayers before and after diamond deposition. Field Emission Scanning Electron Microscopy (FE-SEM, LEO Supra 35) and Energy Dispersive X-Ray Spectroscopy (EDS, Oxford Instruments Ltd., model Inca 300) were used to observe the PVD and multilayer (PVD+diamond) films surfaces. In order to check both the CrN and the multilayer (CrN+diamond) wear resistance, tribological tests with alternative dry-sliding motion were performed using an Optimol SRV tribometer at room temperature. Fixed coated samples were tested at 50 N load with back-and-forth sliding (stroke length: 2 mm, frequency: 50 Hz, duration: 30 min) of the upper SAE52100 steel ball (10 mm diameter). A Taylor-Hobson instrument, the Form Talysurf Intra, was used to assess the wear volume of CrN-coated and multilayer (CrN+diamond) coated materials. To measure wear volume, 4000 points per mm along stylus axis and 500 roughness profiles per mm along the normal axis were acquired for each sample. TalyMap software was used for data analysis and image processing.

Fig. 3 shows the morphology of the CrN films after PVD-arc deposition and after 9 h diamond CVD. The CrN films exhibited the typical surface roughness related to the presence of droplets (or “macros”), produced during arc evaporation (Fig. 3a) [37]. After 9 h CVD, the droplets were embedded in the 6-μm thick diamond film and the surface roughness of the multilayer coated material was mainly caused by the morphology of the <100>{111} textured diamond coating (Fig. 3b).

Fig. 4 shows the XRD patterns of the CrN interlayer before and after 9 h CVD are shown. The XRD pattern of the as-deposited CrN film shows only broad peaks attributable to chromium nitride with NaCl structure [38]. After 9 h CVD deposition, besides sharper CrN peaks, peaks attributable to Cr₇C₃ [39] and Cr₃C₂ [40] were detected. This means that chromium nitride was partially transformed into chromium.
carbides. This finding agrees with Buijnsters et al. [41], who studied diamond deposition on steels using arc-plated CrN interlayers. Fig. 4 also shows that sharp and intense (111) and (220) diamond peaks were also present.

Fig. 5 shows the wear groove morphology obtained for substrate (WC-6 wt.% Co), CrN film and multilayer film (CrN+diamond) after the 30 min dry sliding wear test. An extended worn zone, characterized by a rather tiny average depth (0.4 μm), was detected on WC-Co sample. This wear track was the largest one measured. Its corresponding wear volume was 1.6 × 10^6 μm^3. The wear behaviour of the CrN film was different. Wear track on CrN coated sample was sharper and deeper, with wear volume of 3.8 × 10^6 μm^3. The average and maximum depths of the wear track were 3.9 and 7.9 μm, respectively. Therefore, recalling that the thickness of the CrN film was 7 μm, these data indicate that the CrN coating was completely worn in the centre of the wear track, and it was this which determined the larger wear volume. However, the greater stiffness of the CrN film reduced the penetration of the steel sphere during the test, and led to a sharper wear track. It is worth noting that, after the 30 min sliding test, no wear of the multilayered CrN+9 h CVD diamond film occurred. Fig. 6 shows the SEM pictures of the multilayered CrN+9 h CVD diamond film after the sliding test. Bright areas in the backscattered electrons image (panel a) represent both steel fragments and oxidized debris. The secondary electrons image (panel b) shows that faceted diamond grains on the sample surface were not worn during the test. These findings clearly demonstrate that CrN PVD film improved diamond film properties. Diamond films deposited on CrN possess excellent wear behaviour, characterised by the absence of any measurable wear volume (after the sliding tests). The good adhesion of diamond on CrN PVD films could be caused by chromium carbide formation on PVD film surfaces during CVD process and this will be the subject of future studies on Cr-base interlayers for the deposition of adherent diamond films on hard metals.

2.3. WC-Co heat treatments

Heat treatments of as-ground WC-Co modify the substrate surface morphology and reduce surface Co concentration. Although the mechanisms that improve adhesion level differ, in most cases some kind of surface reconstruction occurs. Saijo et al. [42] pre-treated WC substrates, previously sintered without any binder, in H₂−O₂ plasma at 910°C. This thermo-chemical treatment led to WC decarburization, which resulted in the formation of a surface layer of metallic tungsten. During the subsequent diamond CVD process, metallic W recarbürized into fine WC grains, thus increasing the diamond/substrate contact area at the interface. An Al−20% Si alloy was turned in order to check the cutting performance of decarburized and non-decarburized coated tools. Decarburized tools that have undergone the thermal treatment in the plasma atmosphere prior to diamond film deposition exhibited a much longer tool life than non-carburized inserts. Zhang et al. [43] used a
multiple chemical treatment including microwave oxidation of WC-6% Co substrates in a CO₂–O₂ plasma at 900 °C. The microwave oxidation led to the formation of CoWO₄ and WO₃ phases at the substrate surface, which were then removed by etching in a heated alkaline solution. Lastly, cleaning treatment in an ultrasonic bath using HCl and diamond powder was finally performed to remove surface Co and enhance nucleation by diamond seeding. The coated tools had considerably longer lives, a fact revealing by boring tests involving both diamond coated and uncoated borers for an Al-based 35 vol.% SiC composite.

Another effective way to eliminate surface Co and to induce the coarsening of superficial WC grains is to resinter WC-Co parts in a protective atmosphere. The thermal treatment performed by Oles et al. [44] produced WC grain growth at the surface, increased the roughness (Rₚ > 0.6 μm) without forming subsurface porosity and eliminated surface Co by evaporation. Diamond films deposited on these heat-treated substrates exhibited excellent machining characteristics. In fact, the diamond coated tools provided greater tool lives than polycrystalline diamond (PCD) tools in turning Al–20% SiC [44].

Singh et al. [45] developed a multiple-pulse laser irradiation technique to create a microrough surface morphology. The surface modification was induced by multiple laser pulses having irradiation densities ranging from 3.0 to 6.0 J/cm², i.e., near the ablation threshold. However, further treatment with HCl solution was necessary to remove surface cobalt prior to diamond deposition by HFCVD. As assessed by Rockwell indentation tests, the adhesion of 5-μm thick films deposited on laser-treated WC-10 wt.% Co substrates was excellent. However, the authors did not provide any machining data.

Surface modification of WC-Co inserts can also be induced by exposing the material to a monohydrogen rich atmosphere in the very same HFCVD reactor used for diamond deposition [46]. The gas phase can be activated by the hot filament to induce the thermal dissociation of molecular H₂ in the presence of suitable amounts of hydrocarbons. In fact, small amounts of carbon must be present in the gas phase to suppress WC-Co decarburization. As a result of this undesired, brittle γ-phases (Co₃W₄C) would form in agreement with the following reaction [47]:

\[ x\text{WC} + y\text{Co} + 2(1-x)\text{H}_2 \rightarrow (1-x)\text{CH}_4 + \text{Co}_x\text{W}_y\text{C}. \]

To promote surface reconstruction and Co evaporation, the substrate temperature has to be around 1000 °C. Fig. 7 shows the WC(100) diffraction peaks (Cu Kα radiation, λ = 1.54178 Å) of an as-ground 3-μm sized WC-6%Co insert and the same sample after in situ thermal pretreatment (T-pretreatment). The pretreatment was performed by exposing the sample in the HFCVD chamber, for 3 h, to a gas phase (0.5% CH₄/H₂, 4.8 kPa total pressure) activated by a hot Ta filament (filament temperature: 2300 °C, filament-to-substrate distance: 4 mm, substrate temperature: 1000 °C). At this point it is important to recall that the sampling depth of the XRD technique was less than 2 μm, as calculated from the mass absorption coefficient of WC-Co and the optical path of incident and diffracted beams [48]. The broad WC diffraction peaks of as-ground hard metal substrates can be ascribed to lattice disorder, damage and microstrain induced by the grinding process [48,49]. After the T-pretreatment, the full width at half maximum (FWHM) of the WC (100) peak decreased from 0.35° to 0.22°. This finding, combined with SEM observation of the sample (Fig. 8), shows that recrystallization of the outermost layer of WC, and consequent surface corrugation, occurred. The heavily strained as-ground cemented carbide underwent nucleation and growth of new tungsten carbide crystallites, whose size was smaller (0.5–2 μm) than the average size of bulk WC grains (Fig. 8). This surface WC recrystallization must result from local high temperatures caused by both H-atoms recombination at the substrate surface and release of the stored strain energy. Both these effects could determine higher surface temperatures than those measured by the thermocouple (1000 °C).

3. Cutting performance of CVD diamond

3.1. Effect of substrate pretreatments

To check the performance of CVD diamond coated tools, turning tests were performed on metal matrix composites...
(MMCs) using triangular cutting inserts (TPGN 160308). Two sets of Co-cemented tungsten carbide (WC-6 wt.% Co) cutting inserts (supplied by FILMS S.p.A., Italy) with fine (1 μm), and medium (3 μm) average grain size were used as substrates. Table 1 provides the nominal chemical compositions, density, hardness and coercivity of the sintered parts.

The as-ground substrates were rinsed with acetone and ethanol. Since adhesion strength can be improved by increasing the interfacial area between the film and the substrate [50], WC-Co inserts were thermally or chemically pretreated.

Since diamond films deposited on thermally pretreated fine- and coarse-grained substrates do not ensure satisfactory adhesion levels [46], only 3-μm grained WC-Co substrates were submitted to thermal (T) pretreatment and diamond CVD.

The T-pretreatment was composed of an initial 3 h heat treatment in the Hot Filament CVD reactor, under typical CVD conditions, using a 0.5% CH₄/H₂ atmosphere and 1000 °C substrate temperature. Total chamber pressure was 18 Torr (2.4 kPa) and the gas phase was activated using a tantalum filament heated to 2350 °C. No diamond crystals nucleated at the WC-Co surface because of the low concentration of methane and the high substrate temperature [51]. However, as described in § 2.3 and shown in Fig. 8, surface reconstruction occurred. Following the heat treatment, Caro’s acid (H₂O₂ + H₂SO₄) was used to etch the inserts for 10 s in order to get a surface Co concentration value below the EDS technique detection limit.

The chemical etching (M-pretreatment) involved the two-step procedure described in § 2.1. The first etching was performed using Murakami’s reagent (10 g K₃[Fe(CN)₆] + 10 g KOH + 100 ml of water, 10 min etching time in ultrasonic vessel). The second etching step was performed using Caro’s acid to wash surface Co out. The pre-treated substrates were labeled with a number (1 or 3), referring to the average grain size, and a capital letter according to the pretreatment (M or T). They were then submitted to diamond CVD. Process conditions ensuring 1.5 μm/h deposition rates were employed (1.5% CH₄ in H₂, 2300 °C filament temperature). Substrate temperature was approximately 750–780 °C and deposition time 16 h. The CVD coated inserts are listed in Table 2, as well as the thickness of the diamond coatings measured by IR interferometry with a Biorad-Digilab FTS40A spectrometer. Fig. 9 shows the cutting edge of a typical triangular cutting insert coated by CVD diamond.

The inserts were used to machine Al–10% Al₂O₃ MMC bars (W6A.10A.T6 by Duralcan). Dry turning tests were performed using a SAG 210 Graziano machine and the cutting conditions were the following: cutting speed, \( v_c = 450 \text{ m/min} \); feed rate, \( f = 0.4 \text{ mm/rev} \); depth of cut, \( DOC = 1.5 \text{ mm} \). The flank wear of the inserts was measured using SEM at selected time intervals. For the purposes of comparison, tests were also performed on PCD (Polydiam 4 by Cafro S.r.l., Italy) and TiN coated inserts (by Sandvik) possessing the same geometry.

Fig. 10 shows the flank wear of PCD, CVD- and TiN-coated inserts as a function of cutting time. It is worth noting that when diamond was deposited by CVD on both fine- and medium-grained M-pretreated substrates (1M CVD and 3M CVD tools), coated inserts exhibited a flank wear rate exceeding 0.2 mm flank wear rate after about 3 min of machining. This is comparable to PCD inserts and much lower than TiN-coated inserts. Very good wear resistance was also shown by thermally pretreated diamond coated tools (3T samples), whose excellent behaviour confirmed that suitable selection of WC-Co microstructure and pretreatment produces high-performance diamond coated cutting tools.

### 3.2. Effect of film microstructure

Another important issue concerning the abrasive wear of diamond coated tools is film microstructure. Kamiya et al. [52] showed that the grain size of chemically vapour deposited diamond influences both fracture and adhesive toughness of the film. High coating fracture toughness values would enhance the tool life, being the abrasive wear resistance proportional to \( K_{IC}^{3/4} H^{1/2} \), where \( K_{IC} \) is the fracture toughness and \( H \) the hardness of the film [53].

In order to evaluate the role of CVD diamond microstructure on the cutting performance of coated inserts, diamond films were deposited onto fine-grained hard metal (WC-6 wt.% Co) triangular inserts (TPGN 160304) previously submitted to 5 min etching using Murakami’s and 60 s etching in HNO₃ + H₂O₂ solution. Diamond films (thickness 5 μm) were then deposited using (i) constant CH₄ concentration (1.0%) and (ii) modulated CH₄ concentration at 3.0% and 1.0% for 10 and 40 min, respectively [54]. Constant concentration of methane in the CVD chamber, during diamond deposition is the conventional approach to deposit diamond films. The timed CH₄ concentrations are an integral part of the time-modulated CVD (TMCVD) process which was proposed by Q. H. Fan et al. [55]. This allows the enhancement of the secondary

### Table 1

<table>
<thead>
<tr>
<th>WC-Co grain size [μm]</th>
<th>Nominal composition (wt.%)</th>
<th>Density (g/cm³)</th>
<th>Hardness (HV10)</th>
<th>Transverse strength (MPa)</th>
<th>Coercivity (Oe)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>WC-5.8% Co-0.2% VC-0.2% TaC</td>
<td>14.91</td>
<td>1740</td>
<td>3350</td>
<td>275</td>
</tr>
<tr>
<td>3</td>
<td>WC-6% Co</td>
<td>14.96</td>
<td>1445</td>
<td>3538</td>
<td>135</td>
</tr>
</tbody>
</table>

### Table 2

<table>
<thead>
<tr>
<th>Insert Substrate grain size</th>
<th>Substrate pretreatment</th>
<th>Deposition time</th>
<th>Film thickness</th>
</tr>
</thead>
<tbody>
<tr>
<td>1M 1 μm M</td>
<td>16 h</td>
<td>23 ± 2 μm</td>
<td></td>
</tr>
<tr>
<td>3M 3 μm M</td>
<td>16 h</td>
<td>23 ± 2 μm</td>
<td></td>
</tr>
<tr>
<td>3T 3 μm T</td>
<td>16 h</td>
<td>25 ± 3 μm</td>
<td></td>
</tr>
</tbody>
</table>

(a) M: 10 min etching with Murakami’s reagent +10 s etching with H₂O₂+H₂SO₄; T: in situ 3 h thermal treatment at 1000 °C in monohydrogen-rich atmosphere.
nucleation of diamond, thus refining the coating grain size without any significant deterioration of the Raman quality of the deposited diamond.

Dry turning tests (feed rate 0.05 mm/rev, depth of the cut 0.3 mm, cutting speed 400 m/min) using commercial graphite (ISO 88) as machining material were used to investigate the cutting properties of the inserts. ISO 88 graphite is a highly abrasive material generally used to machine electrodes for electro discharge machines (EDM). The inserts were mounted onto a CNC Kinsbury MHP 50 machine (CTFP 90° tool holder).

Cutting was interrupted at regular time intervals in order to measure the flank wear with an optical microscope (Mitutoyo TM-500). For sake of comparison, machining tests were carried out under identical conditions using uncoated inserts. Each test was repeated twice for each machining condition. Fig. 11 shows the flank wear time evolution of WC-Co, TMCVD- and HFCVD-coated inserts in dry turning of graphite. The results indicated that WC-Co inserts underwent a much faster flank wear than diamond coated tools. However, it is worth noting that conventional HFCVD- and TMCVD-coated tools exhibited different behaviour. In fact, the wear rate of the TMCVD coatings was systematically lower than the conventional diamond coatings. This fact clearly confirms the importance of the coating microstructure in determining the wear behaviour and the overall performance of diamond coated inserts.

3.3. Cutting forces in milling

Most papers published in the relevant literature use turning to evaluate the cutting behaviour of diamond coated tools so few drilling and milling tests are reported [56,57], and tool performance is mainly evaluated by measuring flank wear vs. cutting time or cutting length. No information concerning milling cutting forces is available, although cutting forces have been measured for drilling [56] and MMCs turning operations [58].

Several workpiece materials were considered in order to evaluate the effect of the diamond coating and its surface roughness on the cutting conditions. Material choice was optimised so that a wide range of possible behaviours under cutting conditions was obtained. In this way, a Co–Cr–Mo dental alloy (Biosil® F by Degussa) and a brittle homogeneous material (a clay brick) were employed. The mills were supplied by Metrodent Ltd., UK (HPTX 23 F). All the mills had a 44-mm total length with 32-mm-long cylindrical shank and 12-mm-long conical cutting surface.

Fig. 9. Cutting edge of a WC-Co turning insert coated with 23 μm CVD diamond.

Fig. 10. Flank wear time evolution of PCD, CVD- and TiN-coated WC-Co inserts in dry turning of Al–10% Al2O3 MMC (depth of cut 1.5 mm; feed rate 0.4 mm/rev; cutting speed 450 m/min).

Fig. 11. Flank wear time evolution of WC-Co, TMCVD-, HFCVD-coated inserts in dry turning of graphite (depth of cut 0.3 mm; feed rate 0.05 mm/rev; cutting speed 400 m/min) [54].

Fig. 12. Comparison between uncoated and diamond coated mills after machining Co–Cr–Mo alloy. The uncoated mill underwent quite evident and severe wear [59].
Co–Cr–Mo alloy was submitted to lateral milling (cutting speed 20,000 r.p.m., feed rate 0.01 m/min, depth of cut 3 mm and width of cut 0.5 mm) using an uncoated mill and a mill coated with a 15±2-µm-thick diamond film deposited using process conditions described elsewhere [59]. Prior to HFCVD, the mills were submitted to the M-pretreatment described in § 3.1. Fig. 12 shows the two mills at the end of testing. The catastrophic failure of the uncoated tool is evident. Fig. 13 compares the mean value of the feed force for coated and uncoated mills during milling. It is clear that lower cutting forces were necessary for the coated mill, due to the lower diamond friction coefficient. For both mills, an initial force increase developed at the beginning of cutting after which it levelled out. The uncoated mill failure appeared during this plateau. Finally, a sudden force increase occurred in the final part of the milling pass, due to mill loading (i.e., tool cavities filling by workpiece material). Therefore, although the coated mill exhibited negligible wear at the end of the test, the increase of the feed force clearly shows the importance of chip evacuation. Since chip evacuation can be inhibited by surface roughness of the diamond coating, it was decided the effect of the coating surface roughness on the cutting conditions should be evaluated. So, three HPTX 23 F mills were compared in machining brittle clay bricks, whose chips took the form of a fine powder that tend to pack inside the cut region. The first mill was uncoated, the second mill was coated with a thick diamond coating (15 µm) and the third mill presented a smoother and thinner (5 µm) diamond coating. Prior to HFCVD, the mills were submitted to the M-pretreatment. To obtain a smoother and thinner continuous diamond film, the third mill was also subjected to seeding with 1/4 µm diamond suspension. Fig. 14 shows the different surface morphology of the diamond films deposited on seeded and unseeded mills for 4 h and 12 h, respectively. Surface roughness was measured on the non-sculptured part of the mills on four different paths angularly spaced at 90 °C along the mill circumference. The lowest average roughness was observed on the uncoated mill (Ra=0.029 µm), the smoother diamond coating exhibited an average roughness of 0.81 µm, while considerable roughness was measured on the mill with the thicker coating (Ra=3.0 µm).

Central milling tests were performed on a column-and-knee type milling machine. A high cutting rate was used (20,000 rpm, feed rate 76 mm/min, depth of cut 6 mm).

Fig. 15 shows that the feed force for the mill with the smoother coating was lower than that measured in the case of the uncoated mill for all the cut length. The increase of the feed force for the uncoated mill with time makes it possible to predict a failure for thermal degradation. It is worth noting that, due to the effect of bad chip removal connected to the excessive roughness, the mill with thicker and rougher diamond coating broke after a short cut length due to the development of excessive cutting force and consequent flexure of the shank.

![Fig. 13. Comparison between the cutting forces of uncoated (grey line) and diamond coated (black line) mills during lateral milling of Co–Cr–Mo alloy [59].](image1)

![Fig. 14. SEM micrographs of 15-µm-thick (panel a) and 5-µm-thick (panel b) diamond coatings on WC-Co mills. Prior to deposition of the thinner diamond film (panel b), the substrate was seeded with 1/4 µm diamond suspension in order to get a thin, continuous diamond coating.](image2)

![Fig. 15. Comparison between the cutting forces during milling of clay bricks using an uncoated WC-Co mill, and cutting forces in mills coated with smooth and rough CVD diamond films.](image3)
4. Conclusions

Diamond coated hard metal (WC-Co) tools exhibit wear resistance comparable to or even better than polycrystalline diamond tools (PCD). The pretreatment of the substrate plays an important role in determining the adhesion of the diamond coating onto hard metal and the final properties of coated parts. Thermal treatments, which induce surface reconstruction of the carbide, have been shown to be as effective as chemical etching pre-treatments. However, the effectiveness of a given pretreatment can depend on the carbide microstructure, and, as a general rule, coarse-grained WC-Co substrates do not exhibit mechanical properties (such as hardness and transverse rupture strength) which match the excellent wear resistance of CVD diamond in a satisfactory way.

Although diamond adhesion represents a critical issue, the overall tribological performance of diamond coated tools depends on coating microstructure and smoothness. In fact, grain size might influence coating toughness, and a smoother film can facilitate chip evacuation and then reduce cutting forces. Therefore, appropriate selection of substrate pretreatment and coating microstructure and roughness would allow CVD diamond to gain recognition as a material with the potential to enhance performance and productivity in an ever-broadening array of industrial applications.

Acknowledgements

The Author wishes to thank everyone with whom he has worked for many years: Prof. E. Traversa, Dr. A. Allegri, Dr. M. Barletta, Dr. S. Guarino and Dr. F. Quadrini (University of Rome Tor Vergata, Italy), Dr. G. Mattei (ISC-CNR, Rome, Italy), Dr. G. Marcheselli (FILMS S.p.A., Italy), Dr. W. Ahmed, Rome Tor Vergata, Italy), Dr. G. Mattei (ISC-CNR, Rome, Italy), Dr. S. Guarino and Dr. F. Quadrini (University of Rome Tor Vergata, Italy).

The financial support of Fabbrica Italiana Leghe Metalliche Sinterizzate (FILMS) S.p.A. is gratefully acknowledged.

References