Hard yet Tough Nanocomposite Coatings – Present Status and Future Trends

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Toughness and hardness are important aspects for coating applications in manufacturing industry. Extensive theoretical and experimental efforts have been made to synthesize and study nanocomposite coatings with super hardness and high toughness. The materials can be hardened through various or combined hardening mechanisms. However, for engineering applications, coating toughness is as important as, if not more than, super hardness. At present, there is neither a standard test procedure, nor a standalone methodology for the assessment of thin film toughness. The determination of the toughness is still a difficult task, and very much a fully open problem. In this article, we review the hardening and toughening mechanisms of nanocomposite films, and the toughness characterization techniques. Based on these reviews, an outlook will be presented in the concluding remarks.

Introduction

Nanostructured or nanocomposite coatings are a new branch of materials that possess unique physical and mechanical properties. A nanocomposite coating comprises of at least two phases: a nanocrystalline phase and an amorphous phase, or two different nanocrystalline phases. These nanocomposite coatings represent a new class of materials, whose mechanical and tribological properties are not subjected to volume mixture rules, but depend on grain boundary effects, and on synergetic interactions of the composite constituents owing to the size effect.¹ Now, it is accepted that materials can be classified as hard, superhard or ultrahard for hardness over 20, 40 or 80 GPa, respectively.² Hard, superhard or even ultrahard nanocomposites have been a hot research topic recently.²⁻¹⁰ Hard coatings are used in many applications, for example, cutting and polishing tools, molds, dies, hard disk and other wear-resistant applications. However, for engineering applications, hardness must be complimented with high toughness, which is a property of equal importance as hardness.¹¹⁻¹³ Therefore, it is vital to master the formation of hard films with high toughness (see Figure 1).

Toughness is an important mechanical property related to the materials resistance against shock loads, and it describes the resistance against the formation of cracks resulting from stress accumulation in the vicinity of structure imperfections. In an energetic context, toughness is the ability of a material to absorb energy during deformation up to fracture.¹¹ According to this definition, toughness encompasses the energy required to create a crack and to enable the crack to propagate until fracture.
Therefore, a high toughness coating has high resistance to formation of cracks under stress, and in the mean time, high energy absorbance to deter crack propagation, whereby preventing chipping, flaking or catastrophic failure. Creating tough, fracture-resistant ceramics has been a focal point of ceramics research for decades. Ceramics have inherently low fracture toughness, and are thus subject to brittle fracture (in contrast with ductile metals, which are much tougher because of the plasticity induced by dislocation motion). It is known that mechanical properties of a solid depend strongly on the density of dislocations, interface-to-volume ratio and grain size.\[14\]

An enhancement in damping capacity of a nanocomposite solid may be associated with grain-boundary sliding,\[15\] or with energy dissipation mechanism localized at interfaces.\[16\] A decrease in the grain size significantly affects the yield strength and hardness. The grain boundary structure, boundary angle, boundary sliding and movement of dislocations are important factors that determine the mechanical properties of the nanocomposites. In most cases, coatings are applied to protect surfaces from the consequences of mechanical loads. A greatly simplified examination of the abrasive wear resistance in a variety of materials as a function of the fracture toughness (as applied to bulk materials) reveals a dependence, namely the maximum wear resistance results from a specific, favorable combination of hardness and toughness.\[14\]

**Design Methods for Hard yet Tough Nanocomposites**

**Hardening Mechanisms in Nanocomposite Coatings**

Hardness is defined as the resistance of a material to plastic deformation. Plastic deformation of crystalline materials occurs predominantly by dislocation movement under applied load. It means that a material with enhanced hardness has a higher resistance to dislocation movement. Very often, several of the classical hardening or strengthening mechanisms are active in hard coatings deposited by plasma-assisted vapor deposition techniques. Such hardening mechanisms work by providing obstacles for the dislocation movements, which can also be applied to some extent to hard films. Hindering of dislocation movements can be achieved by: (i) grain boundary hardening, (ii) solid solution hardening, (iii) age hardening and (iv) compressive stress hardening.

**Grain Boundary Hardening**

In a recent review,\[18\] Veprek has analyzed the design criteria to produce superhard or even ultrahard nanocomposite coatings. From material-selection point of view, combinations of nanocrystalline transition-metal nitrides, for example TiN, W2N and VN, with amorphous Si3N4 or BN as the grain boundary phase, have a potential to achieve coating hardness exceeding 40–50 GPa. This hardening behavior, especially in the regime where the crystallite size \(d\) is less than 10 nm, is because the tiny crystallites will restrict the grain boundary sliding in case of a thin amorphous grain boundary.\[2\] One of the best known theories based on dislocation pile-up is described by the Hall-Petch equation.\[19,20\] The relationship indicates that the hardness of the material is inversely proportional to the square root of the grain size. However, as grain sizes are reduced to the nanometer scale, and the percentage of grain boundary atoms increases correspondingly, this traditional view of dislocation-driven plasticity in polycrystalline metals needs to be reconsidered. In a sample with grain diameters of 20 nm, 10% of atoms are located at grain boundaries. Dislocation sources and pileups are hardly expected to exist in such a material, and deformation is believed to be carried mostly by the grain boundaries via a certain accommodation mechanism.\[21\]

For very small grains, the numbers of atoms at inter-granular boundaries and inside grains become comparable. In this case, softening may happen via inter-granular slips, as the inter-granular boundaries cease to be barriers for the motion of dislocations.\[22\] The search for very hard materials is coupled with the study of low-compressibility solids, which have high values of the bulk \(K_b\) or shear \(G\) modulus. Transition-metals-containing compounds have large cohesive energies and high \(K_b\), associated with the distribution of their valence electrons between bonding and anti-bonding states within the partly filled electronic bands. Transition metal carbides, nitrides, borides, or oxides have very low compressibility and also often possess high hardness. Based on thermodynamically driven spinodal segregation in a binary or ternary system, Veprek et al. proposed a

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**Figure 1.** Classification of nanocomposites according to their hardness and toughness.\[14\]

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\[d\] represents the grain size.
design to achieve superhard or even ultrahard coatings; a highly thermal stable nanocomposite structure with ≤10 nm size crystallites of the transition metal nitride separated by about one monolayer of Si3N4 was synthesized (see ref.[2] and references therein). The approximately one monolayer thin tissue of Si3N4 acts as a ‘glue’ for the MeN nanocrystals to achieve ultra-hardness, thus avoiding the reverse Hall-Petch effect. Though the original value given was 103 GPa, more recently studies show that the Fischerscope used normally gives too high hardness values for superhard coatings.[23] Nevertheless, the design concept should be on the right track. Subsequently, many superhard nanocomposites with similar design have been synthesized, but none approaches the hardness of diamond. Figure 2 shows the hardness of nc-TiN/a-Si3N4 coatings deposited by close-field unbalanced magnetron sputtering as a function of the silicon nitride fraction. Silicon-free TiN forms elongated crystallites that are several hundreds of nanometers long and tens of nanometers wide (inset A in Figure 2).[24,25] TiN crystallites become very small at increased Si3N4 contents. At 15–20 at.-% Si3N4, the mean grain size of TiN does not exceed 7 nm; this size is too small for dislocation activities. Therefore, under mechanical loading, such a material can react only by grain-boundary sliding (i.e., by moving single, undeformed TiN nanocrystallites against one another). This process requires more energy than deformation by dislocation movement; hence a higher hardness can be achieved for such coating structures. Estimations of the mean grain separation at this composition show that only a few monolayers of silicon nitride separate the nanocrystallites (inset B in Figure 2). At high silicon nitride fractions, the mean grain separation becomes so wide, that ordinary crack propagation in Si3N4 takes place, and the hardness approaches that of Si3N4 (inset C in Figure 2).

### Solid Solution Hardening

Solid solution hardening comes from the lattice distortion as a result of insertion of atoms of an alloying element in the interstitial location or substation for some of the host atoms. This is, perhaps, one of the oldest hardening methods used in bulk materials. In thin films or coatings, the same principle works too. TiCN,[26] TiAIN,[27] CrAlN,[28] and CrZrN[29] are a few popular examples. Formation of a non-equilibrium supersaturated solid solution of B in TiN may contribute to further strengthening of the material[30] since the gliding of the dislocations eventually formed inside the crystallites is hindered by the strain exerted by the insertion of B. To increase the solubility of B in TiN, non-equilibrium growth processes, such as physical vapor deposition (PVD), or plasma-assisted chemical vapor deposition (PACVD) have been used, where a maximum B solubility of 17.4 at.-% has been reported for PACVD[30] with hardness up to 43 GPa. For reactive arc evaporation from Ti-B compound targets, the highly ionized flux of film-forming species was utilized to synthesize a promising nanocrystalline metastable supersaturated solid solution of B in TiN at lower nitrogen fractions.[31] The maximum in hardness (34.5 GPa) is obtained when the crystallites are 6–8 nm in size, and are strained by the formation of a substitutional solid solution of B in TiN. The formation of a metastable solid solution was confirmed by XRD-observed widening of the lattice upon B substitution of N as compared to TiN. Therefore, the mobility of deformation-induced dislocations is hindered in these small crystallites that are initially dislocation-free. At high N2 fractions, the substitutional B sites in the TiN crystallites are replaced with additional N to form energetically favorable unstrained TiN crystals. This unstrained TiN crystals combined with high fraction

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**Figure 2.** Hardness of nc-TiN/a-Si3N4 as a function of silicon nitride fraction.[24,25] Insets illustrate the schematic nanostructure for different compositions.

**Figure 3.** Indentation hardness and reduced modulus values for Ti-B-N coatings for different N2 fractions.[31]
of less strongly bonded BN amorphous phase lead to a hardness drop (Figure 3). Another example is the superhard Ti-Al-Si-N film (maximal hardness \( \approx 55 \) GPa) synthesized by a hybrid coating system of arc ion plating and sputtering method.[32] Based on XPS and XRD results, it was suggested that the Ti-Al-Si-N films were a composite consisting of solid-solution \((Ti,Al,Si)N\) crystallites and amorphous \(Si_3N_4\).

Age Hardening

Age hardening is yet another “aged” hardening mechanism in strengthening bulk materials (for example Al alloys and steels). Age hardening achieves higher hardness through precipitation over time of a small and uniformly distributed strengthening phase in the form of particles from a supersaturated solid solution. This works equally well in hard coatings, as demonstrated in ref.[33–35] Supersaturated phases can be easily obtained by non-equilibrium deposition techniques such as PVD or PACVD due to limited atomic kinetics during deposition. As-deposited \((TiAl)N\)[34] and \((TiBN)\)[35] films show a dense columnar microstructure of a supersaturated NaCl structured phase based on fcc TiN whose Ti location is partly substituted by Al to form \((TiAl)N\), and N is partly substituted by B to form Ti(BN). During annealing, \((TiAl)N\) and Ti(BN) coatings undergo spinodal decomposition and transform into coherent cubic-phase nanometer-size domains, resulting in hardness increase.

Compressive Stress Hardening

Ion bombardment during deposition at low temperatures has been utilized to increase the density and modify the morphology of the films.[36] During growth, stresses are generally induced by energetic particle bombardment. The lattice defect arrangements induced by impinging energetic ions are responsible for the stress. As the incoming ion or knock-on atoms possess enough kinetic energy, they knock atoms out of their lattice positions and create secondary collisions. In this way, collision cascades. The atom first knocked absorbs most of the energy. The resulting strong atomic motion along the trajectory of the ions leads to a rearrangement of the lattice atoms.[37] The hardness enhancement results from a complex synergistic effect of the decrease of crystallite size, densification of grain boundaries, and compressive stress upon ion bombardment. Musil et al. reported enhancement of hardness of TiN (up to 80 GPa) and \((TiAlVN)\) (up to 100 GPa) during deposition by means of unbalanced magnetron sputtering at negative substrate bias[38] where a correlation between the hardness enhancement and biaxial compressive stress exists.[39] However, upon annealing at a certain temperature, the hardness may decrease back to “normal” when the induced defects are annealed and the compressive stress relaxes.

Toughening Mechanisms in Nanocomposite Coatings

In order to obtain high hardness in nanocomposite coatings, usually plastic deformation is designed to be prohibited, and dislocation movement and grain boundary sliding are prevented, thus causing a loss in ductility. Ductility is related to toughness, which is very important for hard coatings to avoid catastrophic failure. Veprek[40] indented a 6–10 \(\mu\)m-thin coating on a soft steel substrate with a high load of 1 000 mN and found no cracks. This, however, does not necessarily mean it has high “fracture toughness”: the classical definition of “fracture toughness” measures the resistance to crack propagation – a crack exists first, then how difficult it is to make the crack propagate. The coating in the above example is ‘damage tolerant’ owing to the difficulty in generating a crack, but once a crack is generated, it might propagate in a catastrophic manner and thus result in a brittle failure if it is not truly “tough”. In bulk ceramics, various toughening methods are available: ductile phase toughening, fiber and whisker toughening, transformation toughening, microcrack toughening, etc.[41] In hard coatings, however, investigation on toughening mechanisms lags far behind.

Ductile Phase Toughening

In order to toughen hard ceramic films, a straightforward method is to incorporate a ductile phase. Ductile phase toughening arises from two major mechanisms: the relaxation of the strain field around the crack tip through the ductile phase, and the yielding and bridging of cracks by ligaments of the ductile phase. These two mechanisms can increase the work for plastic deformation. Zhang et al.[42] doped Al into a-C films by co-sputtering of graphite and Al. The composite films exhibit a reduction in toughness from 31.5 to 8.8 GPa. In order to restore the hardness, nanocrystalline TiC (or nc-TiC) was embedded into the a-C matrix doped with Al to form nc-TiC/a-C(Al) nanocomposite film. The hardness is restored to \(\approx 20\) GPa, while toughness remained high (indentation plasticity of 55%) and residual stress low (only 0.5 GPa).[43] Although toughness is not measured due to lack of proper measurement facilities for such a high toughness coating, the nc-TiC/a-C(Al) nanocomposite film is evidently very tough: the optical comparison of scanning scratch tracks between nc-TiC/a-C (Figure 4a) film and nc-TiC/a-C(Al) (Figure 4b) shows that without doping of Al, the coating basically fails in a brittle way (Figure 4a); with doping of Al, however, the scratching tip ploughs into the coating causing the scanning amplitude to reduce along the scratch direction (Figure 4b). Co-sputtering of Ti, TiNi
and Si₃N₄ targets produced nanocomposite coatings where the amorphous metallic Ni increased the toughness of nc-TiN/a-SiNₓ coating. Doping from 0 to 40 at.% Ni brings about an increase in toughness from 1.15 to 2.60 MPa·m¹/₂, but at the expense of hardness, which diminishes from 30 to 14 GPa.

Phase Transformation Toughening

Some phase transformation occurs under stress and is accompanied by absorption of a large quantity of energy, which could be used to toughen a material. Transformation from tetragonal to the monoclinic phase takes place under stress. The process occurs in the stress field around the tip of the crack and is accompanied by a volume increase of about 4%. The resultant strain during the transformation relieves the stress field and absorbs the fracture energy, whereby toughens the material. One example is ZrO₂-toughened ZrB₂ composite. In order to facilitate the phase transformation, retention of the high-temperature tetragonal phase is key, which can be easily realized during deposition. Sprio et al. deposited thin zirconium oxide films by radiofrequency (RF) magnetron sputtering. The films have a mixture of tetragonal and monoclinic zirconia phases. It was shown that increasing the substrate bias power disrupted the columnar grain growth. TEM confirmed a reduction in the intergranular porosity, but also an increase in lateral defects. These defects are hypothesized to be stress-induced microcracks caused by a tetragonal to monoclinic phase transformation. Tetrahedral zirconia has a high toughness and is proposed to increase properties of brittle substrates. Trinh et al. prepared γ-Al₂O₃-ZrO₂ thin films by RF magnetron sputtering with hardness up to 30 GPa. However, owing to the lack of proper toughness measurement instrumentation, toughness was not measured.

Compressive Stress Toughening

Generally cracking is initiated by tensile stress, thus compressive residual stress has to be overcome first before a crack is initiated in tension. When a coating has a high compressive residual stress to start with, the coating will be able to take more tensile strain before fracture, as is demonstrated in Zr-Cu-O coatings (see Figure 5). Although a certain level of compressive stress can increase toughness, excessively large compressive stress can cause film delamination or cracking.

Nanotube Toughening

The discovery of carbon nanotube (CNTs) has opened up new avenues in producing unique carbon-based materials. Novel mechanical tests on individual CNTs and atomistic calculations suggest that CNTs have ultra-high elastic modulus approaching 1 TPa, and exceptional tensile strengths, in the range of 20–100 GPa. Due to their outstanding properties, carbon nanotubes have attracted a growing interest and are considered to be the most promising materials for applications in nanoeengineering. Toughness enhancement can be obtained through crack deflection, crack bridging and CNT pull-out. CNT toughened bulk ceramics are attractive, and outstanding results have been successfully achieved, for instance, in Al₂O₃ matrix and...
BaTiO$_3$ matrix. However, nanotube toughened thin films is yet to be realized, due to difficulties in controlled alignment, interface reaction and reasonable volume fractions. Xia et al.$^{[63,67]}$ anodized aluminum in order to obtain an amorphous alumina matrix having pores with diameters in the 30–40 nm range. Multiwall carbon nanotubes were grown by CVD in this matrix to realize CT/alumina composite layers of 20 to 90 $\mu$m in thickness. Crack deflection at the CNT/matrix interface (Figure 6a), crack bridging by CNT (Figure 6b), CNT pull-out (Figure 6c) are all observed.

Composition or Structure Grading Toughening

Graded inter-layering is a known effective method to reduce cracks concentration and enhance adhesion between coating and substrate. In the gradient design, the substrate is first coated with a high adhesion layer, and then the coating constituents are allowed to vary homogeneously or heterogeneously while the coating thickness builds up. With biased graded (varying substrate bias voltage while coating thickness builds up) in magnetron sputtering, Zhang et al.$^{[68]}$ prepared a 1.5 $\mu$m a-C gradient coating on tool steels with moderately high hardness (25 GPa), but very high toughness (plasticity 57.6%). During deposition, substrate bias gradually increased from –20 to –150 V, and a graded sp$^2$/sp$^3$ was achieved through coating thickness. The bottom layer has the highest sp$^2$ fraction for high adhesion, whereas the top layer possess the highest sp$^3$ fraction to render high hardness. Bias-graded coating design enhances the toughness and the adhesion of the coating on tool steel (Figure 7). The adhesion strength increased more than two times as compared to the same coating deposited at constant bias (–150 V). Pei et al.$^{[69,70]}$ changed the substrate bias in order to change the microstructure of the nc-TiC/a-C:H coatings prepared by close field unbalanced magnetron sputtering. Increasing the bias resulted in a clear transition from columnar to glassy microstructure in the coating (Figure 8). Substrate bias also simultaneously and greatly enhanced the coating hardness, toughness and adhesive strength. The coating with glassy microstructure exhibits
a substantial toughening. However, a word of caution on the actual toughness value given in the paper— the value of few tens of MPa·m\(^{1/2}\) is unbelievably high, as also pointed out by the authors, probably due to experimental limitations. Ion bombardment can densify the structure and prevent the formation of columns which are detrimental in terms of microcrack initiation and propagation under load. The induced residual stresses grow as bias increasing.

**Toughness Characterization**

In a classical definition, toughness measures the ability of a material to resist the crack propagation until fracture. For bulk materials and very thick films, toughness can be easily measured according to ASTM standards.\(^{71,72}\) For thin films, there is no proper definition for toughness, since in thin films or coatings, the cracks do not necessarily pre-exist, and thus energy has to be used to initiate the crack. Thin film toughness measurements are difficult or tedious in practice due to the thickness limitations. Although many proposed methods can be used in some cases, a standard procedure or a universally accepted methodology does not exist yet for ceramic thin film or coating toughness measurements. Qualitatively, toughness can be simply estimated by the plasticity\(^{73,74}\), the MDP (microhardness dissipation parameter)\(^{75,76}\), and the scratch crack propagation resistance.\(^{77,78}\) The plasticity or MDP are measured as the ratio of the plastic over the total displacement or work. The scratch crack propagation resistance, or CPR, is determined by measuring the lower critical load, i.e. cohesive failure load \(L_{c1}\) and the upper critical load or adhesion failure load \(L_{c2}\) and then calculated by \(\text{CPR} = \frac{L_{c1}}{L_{c2}}\). The CPR measures the scratch resistance to crack initiation and propagation. More detailed quantitative measurement of toughness via bending, buckling, indentation and tensile methods can be found in ref.\(^{78}\) The most widely method used in the assessment of thin film toughness is indentation. When the stress exceeds a critical value during indentation, a crack or spallation is generated. Failure of the coating is manifested by the formation of a kink or plateau in the load-displacement curve, or by the crack formation in the indent impression. The length, \(c\), of radial cracks is related to the fracture toughness \(K_{IC}\) through the following:\(^{79}\)

\[
K_{IC} = \frac{P}{d} \left( \frac{E}{H} \right)^{1/2} \left( \frac{c}{d} \right)^{3/2}
\]

where \(P\) is the applied indentation load; \(E\) and \(H\) are the elastic modulus and the hardness of the coating, respectively. \(\delta\) is an empirical constant which depends on the geometry of the indenter. Major uncertainties in
this method come from the crack formation threshold, the substrate effect and the crack length measurement.

Another approach based on indentation is through determining the fracture energy. The energy difference before and after through-thickness cracking is believed to be responsible for the coating fracture. The fracture toughness of the film is calculated by:

\[
K_{IC} = \left( \frac{E}{(1 - \nu_f^2)2\pi C_R} \frac{\Delta U}{t^2} \right)^{1/2}
\]

(2)

where \( E \) is the elastic modulus of the coating, \( \nu_f \) is Poisson’s ratio of the coating, \( 2\pi C_R \) the planar crack perimeter, \( t \) the effective coating thickness, and \( \Delta U \) is the strain energy difference responsible for the cracking. This method also inherited the uncertainties for the substrate effect on the deformation fields, the crack size measurement and negligible of the interface debonding.

Tensile testing of free-standing thin films with a pre-crack induced by indentation was recently used to measure the toughness of 0.52 and 1 \( \mu \)m tetrahedral amorphous diamond-like carbon (ta-C).\[^{[83]}\] During preparation of the stand-alone film, wet etching of SiO\(_2\) interlayer between the coating and silicon substrate has been employed. This method has difficulties in obtaining freestanding film without curling, precrack sharpness and accurate measurement of the precrack dimension. A simple two-step tensile method, based on the strain energy difference between two tensile processes is proposed by Zhang et al.\[^{[82]}\] The “two-step” tensile method extends a very elastic substrate coated with the ceramic film in question until the coating fracture. A kink is observed in load-extension curve. The load is then released to zero but re-loaded to the previous extension. The energy difference is considered to be the energy used in creating the fracture. In this method, nc-TiN/a-Si\(_N\) nanocomposite layer with thickness of 3.0 \( \mu \)m gives a toughness value of 2.6 MPa \( \cdot \) m\(^{1/2}\). The method requires very a good elasticity for the substrate (so that when the coating is fractured the substrate is still in the elastic deformation regime), and that the thickness of the substrate to be small enough (for the kink to be visible); plus, there should not be a coating delamination during tension. Currently, there are relatively few toughness data available for thin nanocomposite films due to the difficulties in forming a sensible crack while reducing the substrate effect in the thin films. Toughness values of some nanocomposite are listed in Table 1.

### Future Trends and Directions

For engineering applications, the zest for sole superhardness will subside, while a combination of hardness and toughness will be pursued more by the academics driven by the industrial needs. A sensible route is to toughen superhard coatings in order to achieve what we call “hard yet tough” coatings. In principle, all toughening mechanisms can be applied to the toughening of ceramic thin films or coatings, the real challenge lies in the implementation. In-situ synthesis of carbon nanotube-imbreded nanocomposite (such as nc-TiC/a-C(CNT) or nc-TiN/a-Si\(_3\)N\(_4\)(CNT)) or nanocrystals imbedded in a complete matrix of carbon nanotubes (for instance, nc-TiC/CNT, or even nc-Diamond/CNT) may hold the key (at least, one of the keys) to this ideal “hard yet tough” coating, where the CNT-imbedded matrix or the CNT matrix provides the ultimate “bulk” elasticity and toughness to stop or block crack propagations, and the nanocrystalline phase provides the superior hardness.

In respect to the toughness measurements, the lack of a universally accepted toughness measurement methodology and instrumentation for thin films or coating has effectively hindered the progress of the research. It is hoped that in a not too far future, a universally accepted methodology will be born, and that this will accompany the birth of a toughness measurement apparatus for ceramic thin films or coatings; a stand-alone machine, or

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**Table 1.** Toughness values for some hard coatings.

<table>
<thead>
<tr>
<th>Materials</th>
<th>Toughness MPa $\cdot$ m(^{1/2})</th>
<th>Evaluation Method</th>
<th>Reference</th>
</tr>
</thead>
<tbody>
<tr>
<td>Ti(<em>{x})N(</em>{y})/SiCN</td>
<td>1.0</td>
<td>indentation</td>
<td>^[83]]</td>
</tr>
<tr>
<td>TiN/Si(_{N})</td>
<td>1.15</td>
<td>indentation</td>
<td>^[44]]</td>
</tr>
<tr>
<td>Ni-doped nc-TiN/a-Si(_{N})</td>
<td>2.60</td>
<td>two-step tensile</td>
<td>^[82]]</td>
</tr>
<tr>
<td>YSZ/Au</td>
<td>1.6</td>
<td>indentation</td>
<td>^[84]]</td>
</tr>
<tr>
<td>Nanolayered Cr/a-C</td>
<td>1.81–3.49</td>
<td>indentation</td>
<td>^[85]]</td>
</tr>
<tr>
<td>Ni/Al(_2)O(_3)</td>
<td>2.4</td>
<td>indentation</td>
<td>^[86]]</td>
</tr>
<tr>
<td>ta-C</td>
<td>4.25–4.4</td>
<td>tensile with precrack</td>
<td>^[81]]</td>
</tr>
</tbody>
</table>
perhaps, as an integral part of a hardness testing system. To catalyze the birth of the universal toughness measurement for ceramic thin films or coatings, the community needs first to have a consensus on the definition of the toughness or “fracture toughness” for thin films or coatings (which we believe it should take into account the energy needed to generate the cracks), and regular international seminars should be organized especially on this topic.

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