Kirk-Othmer Encyclopedia of Chemical Technology.

TOOL MATERIALS

Machining of materials using a cutting tool harder than the work material is a common manufacturing operation occurring in the production of a variety of parts. Geometrically defined, single- or multiple-point cutting tools are used to remove the unwanted material from the work material in the form of chips. Generally a numerically controlled (NC) machine tool is employed to provide the required relative motions to produce parts of a given shape, size, and accuracy. A trained operator can produce parts to specifications consistently and economically on a routine basis. Machining processes include turning, drilling, milling, boring, threading, tapping, and broaching. Each operates under a different set of machining conditions. Consequently, the requirements of the tool material differ from one operation to another. The tool materials presented herein are for the most part concerned with cutting operations involving metals and their alloys. Materials used for grinding, polishing, lapping, etc, that use abrasives where the cutting edges are not geometrically defined, ie, have random geometry, and where the geometry changes continuously as the process progresses are outside the scope of this article.

As of the mid-1990s, some estimated $\$300 \times 10^9$/yr was spent on labor and overhead costs alone for machining in the United States (1). This sum does not include the cost of the machine tools and the associated equipment, the cutting tools, the work material, etc. The total cost of the cutting tools used is only a small (ca 1–2%) fraction of this sum and is negligible when compared to the cost of a machine tool. The cutting tool insert, the lowest priced single unit in the machine tool system, however, offers the greatest opportunity for productivity improvement and cost reduction (2).

Machining of metals involves extensive plastic deformation (shear strain of ca 2–8) of the work material in a narrow region ahead of the tool. High tool temperatures (ca 1000°C) and freshly generated, chemically active surfaces (underside of the chip and the machined surface) that interact extensively with the tool material, result in tool wear. There are also high mechanical and thermal stresses (often cyclic) on the tool (3).

1. Modes of Tool Wear

The performance and life of a cutting tool depend on the cutting conditions as well as the combination of tool material, work material, and the lubricant used. Wear on a tool can be in any one of four areas: crater wear on the rake face, flank wear on the clearance face, flank wear on the nose of the tool, and depth-of-cut line (DCL) notch wear in the machining of certain difficult-to-machine materials such as superalloys using ceramic tools (Fig. 1) (5). In addition, part of the tool, eg, the nose, may be deformed plastically owing to inadequate strength at high operating temperatures. Moreover, cracks may be generated on the tool owing to thermal or mechanical cyclic stresses induced during interrupted cutting. A rapid cratering on the rake face of the tool can result either from high temperatures generated at cutting speeds much higher than recommended ones or from high chemical reactivity between the tool material and the work material. Flank wear on the clearance face and on the nose is generally a result of inadequate abrasion resistance of the tool material.

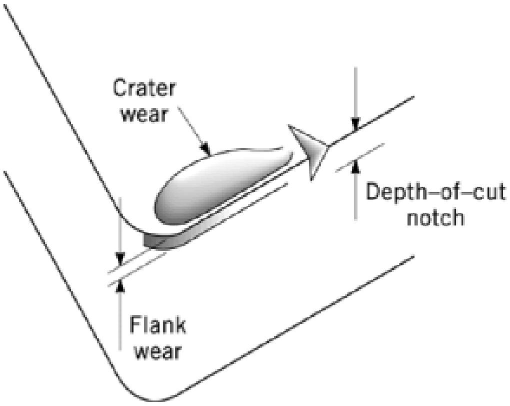

Fig. 1. Schematic showing typical wear modes on a cutting tool.

The rapid cratering of the straight cemented tungsten carbide in machining steels in the late 1930s led to the development of cemented carbide tools containing solid solutions of multicarbide material of W, Ti, and Ta (Nb) for machining steels. Whereas Si_3N_4 tool can machine gray cast iron at very high speed (1500 m/min) with very little wear, and that only on the flank face, the same material when machining steels wears rapidly on the rake face, forming a deep crater. Detailed knowledge of the mechanism of wear, in this case chemical interactions between steel and Si_3N_4, can result in significant improvements in productivity. For example, some success in high speed machining of low carbon steels, malleable cast iron, or nodular cast iron using a ceramic tool containing about 70% Al_2O_3 and the remaining Si_3N_4 and minor amounts of sintering aids have been reported (6). This ceramic tool material was described as a mechanical mixture of Al_2O_3 and Si_3N_4 and not a single phase. The microstructure of Si_3N_4, which consists of elongated grains of β-Si_3N_4 that form an interlocking grain structure, is expected to provide additional toughness to this material. Similarly, SiC whisker-reinforced Al_2O_3 was found to be an excellent tool material for machining nickel-base superalloys, but this material wears rapidly when used in machining steels. Reaction of the micrometer-size SiC whiskers with the steel was postulated because in the case of Al_2O_3–Si_3N_4, the tool did not wear when used in machining steels. Thus where Si is present in the form of Si_3N_4 or SiO_2, wear does not occur, but where Si is present in the form of SiC whiskers the tool wears rapidly. Substitution of other whisker materials might then lead to an appropriate tool for machining steels at high cutting speeds.

2. Tool Materials

A wide range of cutting-tool materials is available. Properties, performance capabilities, and cost vary widely (2, 7). Various steels (see Steel); cast cobalt alloys (see Cobalt and cobalt alloys); cemented, cast, and coated carbides (qv); ceramics (qv), sintered polycrystalline cubic boron nitride (cBN) (see boron compounds) and sintered polycrystalline diamond; thin diamond coatings on cemented carbides and ceramics; and single-crystal natural diamond (see Carbon) are all used as tool materials. Most tool materials used in the 1990s were developed during the twentieth century. The tool materials of the 1990s will likely become the work materials of the twenty-first century.

The properties affecting performance of a cutting tool in machining a given material and a given cutting process can be described as mechanical, thermal, physical, or chemical. Chemical properties control the chemical interaction between the tool, the work material, and the environment. Mechanical properties control the wear, deformation, and fracture resistance. Thermal properties control the heat partition and thermal shock

Table 1. Summary of Properties for Cutting Tool Materials[a]

				Carbides				
Parameter	Carbon and low–medium alloy steels	High speed steels	Cast cobalt alloys	Cemented	Coated	Ceramics	Polycrystalline cBN	Diamond
hot hardness			increasing →					
toughness			decreasing →					
impact strength			decreasing →					
wear resistance			increasing →					
chipping resistance			decreasing →					
cutting speed			increasing →					
depth of cut	light to medium	light to heavy	light to heavy	light to heavy	light to heavy	light to heavy	light to heavy	very light for single crystal diamond
finish obtainable	rough	rough	rough	good	good	very good	very good	excellent
method of processing	wrought	wrought, cast, HIP sintering	cast and HIP sintering	cold pressing and sintering	CVD[b] PVD[c]	cold pressing and sinter-ing or HIP	high pressure–high tempera-ture sintering	high pressure–high tempera-ture sintering
fabrication	machining and grinding	machining and grinding	grinding	grinding or as molded		grinding	grinding and polishing	grinding and polishing
thermal shock resistance				decreasing →				
tool material cost			increasing →					

[a] Overlapping of characteristics exists in many cases. Exceptions to the rule are common. In many classes of tool materials, a wide range of composition and properties are obtainable.

[b] CVD = chemical vapor deposition.

[c] PVD = physical vapor deposition.

resistance of the tool. Thus, the hot hardness determines the abrasion resistance as well as hot deformation resistance. Transverse rupture strength (TRS) determines the toughness of the materials and the ability to withstand the loads applied. Thermal conductivity determines how much of the heat generated at the chip–tool interface is conducted into the tool versus how much goes into the chip. The product $(K \times \mathrm{TRS}/\alpha)$ of thermal conductivity, K, and TRS over the thermal expansion coefficient, α, is termed the thermal shock parameter and determines the tool's ability to withstand the thermal shock experienced during interrupted cutting. The fracture toughness of the tool determines the impact and fracture-resistance of the tool material. The various properties of cutting tool materials are summarized in Table 1.

The cutting tool is an important component of the machining system. Consequently, tool materials significantly affect machining operation productivity. Other elements include cutting conditions, tool geometry, and the characteristics of the work material, nature of parts produced, machine tool, and support system.

The methodology for tool selection is illustrated in Figure 2 (8). Whereas the selection of a particular class of tool material for a given application is relatively simple, selection of a precise tool grade, shape, geometry, chip groove profile, and size is much more difficult. Many times extensive machining tests are conducted in-house before any implementation on the shop floor. General guidelines for the selection of tool materials for different work materials and different machining operations are given in Tables 2 and 3, respectively.

Table 2. Guidelines for Tool Materials

Tool materials[a]	Work materials	Machining operation and cutting-speed range	Modes of tool wear or failure[b]	Limitations
carbon steels	low strength, softer materials, nonferrous alloys, plastics	tapping, drilling, reaming; low speed	buildup, plastic deforma-tion, abrasive wear, microchipping	low hot hardness, limited hardenability and wear resistance, low cutting speed, low-strength materials
low–medium alloy steels	low strength–soft materials, nonferrous alloys, plastics	tapping, drilling, reaming; low speed	buildup, plastic deforma-tion, abrasive wear, microchipping	low hot hardness, limited hardenability and wear resistance, low cutting speed, low-strength materials
HSS and TiN-coated HSS	all materials of low–medium strength and hardness	turning, drilling, milling, broaching; medium speed	flank wear, crater wear	low hot hardness, limited hardenability and wear resistance, low to medium cutting speed, low- to medium-strength materials
cemented carbide	all materials up to medium strength and hardness	turning, boring, drilling, milling, broaching; medium speed	flank wear, crater wear, nose wear thermal, cracks, deformation, fracture	not for low speed because of cold welding of chips and microhipping, not suit-able for low speed application
coated carbides	cast iron, alloy steels, stainless steels, superalloys	turning; medium to high speed, boring, drilling, milling, threading, grooving, parting	flank wear, crater wear nose wear thermal, cracks, deformation, fracture	not for low speed because of cold welding of chips and microchipping, not for titanium alloys, not for nonferrous alloys since the coated grades do not offer additional benefits over uncoated
ceramics	cast iron, Ni-base superalloys, nonferrous alloys, plastics	turning; high speed to very high speed	DCL notching, micro-chipping, gross fracture	low strength and thermo-mechanical fatigue strength, not for low speed operations or interrupted cutting, not for machin-ing Al, Ti alloys
cBN	hardened alloy steels, HSS, Ni-base super-alloys, hardened chill-cast iron, commercially pure nickel	turning, milling; medium to high speed	DCL notching, chipping, oxidation, graphitization	low strength and chemical stability at higher tem-perature, but high strength, hard materials otherwise
diamond	pure copper, pure alu-minum, aluminum-Si alloys, cold-pressed cemented carbides, rock, cement, plastics, glass–epoxy composites, nonferrous alloys, hardened high carbon alloy steels (for bur-nishing only), fibrous composites	turning, milling; high to very high speed	chipping, oxidation, graphitization	low strength and chemical stability at higher tem-perature, not for machin-ing low carbon steels, Co, Ni, Ti, Zr

[a] HSS = high speed steel; cBN = cubic boron nitride.
[b] DCL = depth of cut line.

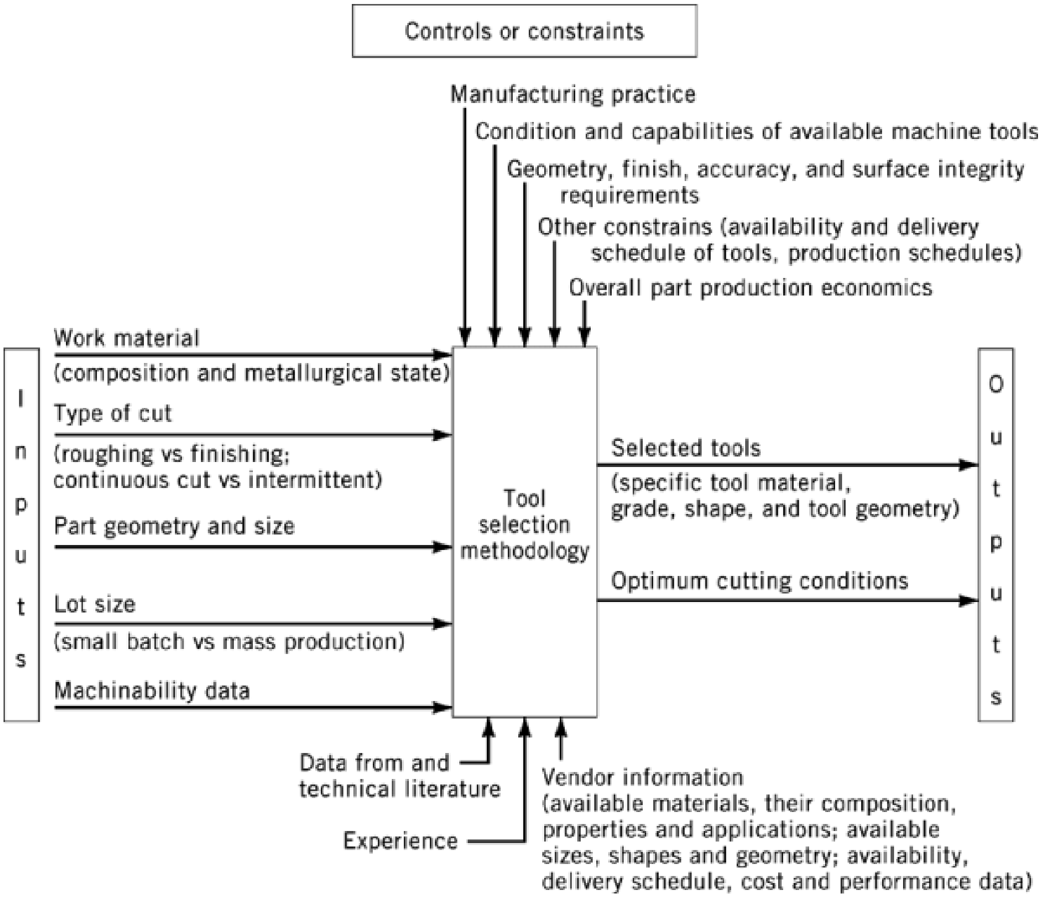

Fig. 2. Methodology for the selection of tool material, grade, shape, size, and geometry, and cutting conditions for a given application (8).

Measurement of hardness (qv) at room temperature is relatively easy; however, it is the hot hardness at the temperature of cutting that is of importance for tool materials. Figure 3 shows the variation of hot (microindentation) hardness of various tool materials measured at different temperatures. The various suppliers of tool materials can be found in References (11–13) and other trade literature.

2.1. Carbon Steels and Low–Medium Alloy Steels

Plain carbon steels, the most common cutting tool materials of the nineteenth century, were replaced by low–medium alloy steels at the turn of that century because of the need for increased machining productivity in many applications. Low–medium carbon steels have since then been largely superseded by other tool materials, except for some low speed applications.

Low–medium alloy steels contain elements such as Mo and Cr for hardenability, and W and Mo for wear resistance (Table 4) (7, 16, 17) (see Steel). These alloy steels, however, lose their hardness rapidly when heated above 150–340°C (see Fig. 3). Furthermore, because of the low volume fraction of hard, refractory carbide phase present in these alloys, their abrasion resistance is limited. Hence, low–medium alloy steels are used

Table 3. Tool Materials for Cutting Operations[a]

Operation	Tool materials[b]	Speed range[c]
single-point turning	low–medium alloy steels, HSS, cemented carbide, coated carbide, ceramics, cBN, diamond	low to very high
drilling	low–medium alloy steels, HSS, solid cemented carbide <2.54 cm	low
tapping	carbon steels, low–medium alloy steels, cemented carbides[d]	low
reaming	HSS, cemented carbides, diamond	low to medium
broaching	HSS, cemented carbide	low to medium
end milling	HSS, solid cemented carbide <2.54 cm, brazed carbides >2.54 cm	low to medium
face milling	HSS, brazed carbides, cemented carbide inserts, diamond, cBN	medium to very high

[a] Refs. 9 and 10.

[b] cBN = cubic boron nitride; HSS = high speed steel.

[c] Low: 30 m/min; medium: 30–150 m/min; high: 150–300 m/min; very high: 300 m/min.

[d] Limited application.

Table 4. Compositions of Carbon and Low–Medium Alloy Steels,[a] and 15

Type[c]	C	Mn	Si	Cr	W	Mo
			Carbon steels[d]			
W1	0.6–1.4					
W2	0.6–1.4[e]					
W3	0.6–1.4			0.5		
			Low–medium alloy steels[f]			
O1	0.9	1.00		0.5	0.5	
O2	0.9	1.60				
O6	1.45		1.00			0.25
O7	1.20			0.75	1.75	0.25

[a] Refs. 8, 14

[b] Remainder Fe in all cases.

[c] W = water – hardening grade.

[d] Available in ranges of 0.1 wt % of carbon content.

[e] Also contains 0.25 wt % V.

[f] Cold worked.

in relatively inexpensive tools for certain low speed cutting applications where the heat generated is not high enough to reduce their hardness significantly.

Low–medium alloy steels are relatively inexpensive and readily available on short notice or for a short run of parts. They can be heat-treated by simple hardening and tempering using relatively inexpensive equipment. They are easily formed and ground, and are processed in many job shops fabricating their own tools. However, these alloys have the following limitations in addition to low hot hardness (see Fig. 3): low wear resistance, poor hardenability, susceptibility to forming quench cracks and grinding cracks, and poor dimensional stability. Choice of a given grade depends on the tool requirement, availability, cost, and other factors.

2.2. High Speed Steels

Toward the latter part of the nineteenth century, a new heat-treatment technique for tool steels was developed in the United States (3, 17) that enabled increased metal removal rates and cutting speeds. This material was termed high speed steel (HSS) because it nearly doubled the then maximum cutting speeds of carbon–low alloy steels. Cemented carbides and ceramics have since surpassed the cutting speed capabilities of HSS by 5–15 times.

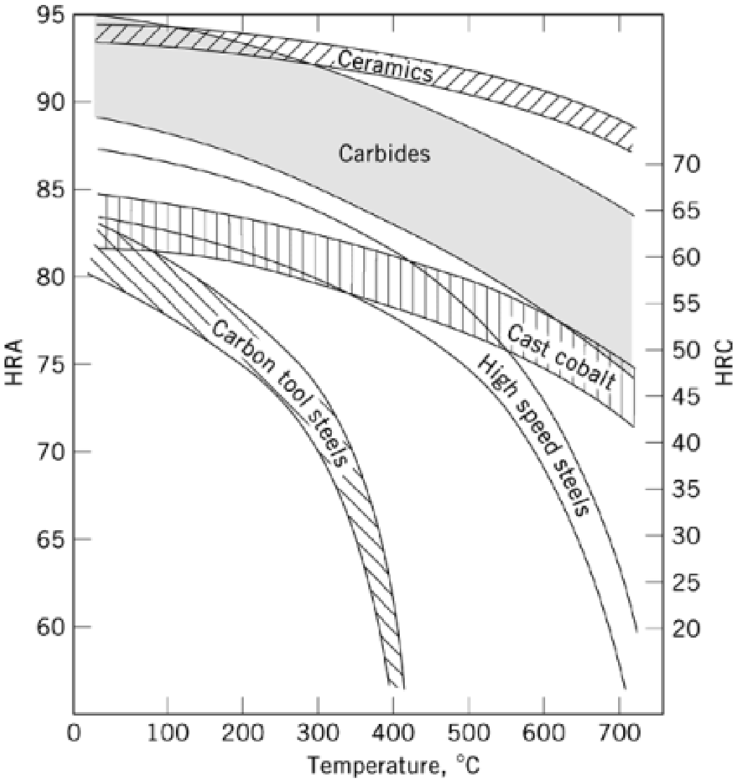

Fig. 3. Hot-hardness of tool materials as a function of temperatures (7). HRA and HRC are Rockwell A and Rockwell C hardness, respectively (see Hardness).

High speed steels contain significant amounts of W, Mo, Co, V, and Cr in addition to Fe and C (18, 19). The presence of these alloying elements strengthens the matrix beyond the tempering temperature, increasing the hot hardness and wear resistance. The materials are readily available at reasonable cost and exhibit the following desirable features: through hardenability; higher hardness than carbon steel and low–medium alloy steels; good wear resistance; high toughness (a feature especially desirable in intermittent cutting); and the ability to alter hardness appropriately by suitable heat treatment. This last facilitates manufacturing complex tools in the soft annealed condition followed by suitable heat treatment for hardening and grinding of tools and cutters to final shape. Associated with the advantages are the following limitations: hardness decreases sharply beyond 540°C, limiting these tools to low speed cutting operations ($<$30 m/min); wear resistance, chemical stability, and propensity to interact chemically with the chip and the machined surface are limited; and the chips tend to adhere to the tool.

Tool steels are broadly classified as T-type or M-type depending on whether W or Mo is the principal alloying element (Table 5), (Fig. 4), (7, 16–19). The original HSS were T-type but concern for the shortage of tungsten, a strategic material, lead to an extensive search for its replacement. Molybdenum was found to serve as an equivalent with additional features in its favor, namely, Mo has half the atomic weight of W, therefore only half as much Mo is required. The two types T- and M- can be used interchangeably because they possess more or less the same properties and have comparable cutting performance. However, M-type steels tend to decarburize more during heat treatment, for which the temperature range is narrow, and hence, care should

Table 5. Chemical Composition of High Speed Steels

	Chemical composition[a], nominal %						
AISI tool steel type	C	Cr	V	W	Mo	Co	W_{eq}[b]
			Tungsten high speed steel				
T1[c]	0.70	4.0	1.0	18.0			18.0
T2[c]	0.85	4.0	2.0	18.0			18.0
T3	1.00	4.0	3.0	18.0	0.60		19.2
T4	0.75	4.0	1.0	18.0	0.60	5.0	19.2
T5	0.80	4.25	1.0	18.0	0.90	8.0	19.8
T6	0.80	4.25	1.5	20.0	0.90	12.0	21.8
T7	0.80	4.0	2.0	14.0			14.0
T8	0.80	4.0	2.0	14.0	0.90	5.0	15.8
T9	1.20	4.0	4.0	18.0			18.0
T15	1.55	4.50	5.0	12.0	0.60	5.0	13.2
			Molybdenum high speed steels				
M1[c]	0.80	4.0	1.00	1.5	8.0		17.5
M2[c]	0.85	4.0	2.00	6.0	5.0		16.0
M3	1.00	4.0	2.75	6.0	5.0		16.0
M4	1.30	4.0	4.00	5.5	4.5		14.5
M6	0.80	4.0	1.50	4.0	5.0	12.0	14.0
M7[c]	1.00	4.0	2.00	1.75	8.75		19.25
M8[d]	0.80	4.0	1.50	5.0	5.0		15.0
M10	0.85	4.0	2.00		8.0		16.0
			High hardness (molybdenum base) cobalt high speed steels				
M30	0.85	4.0	1.25	2.0	8.0	5.0	18.0
M34	0.85	4.0	2.00	2.0	8.0	8.0	18.0
M35	0.85	4.0	2.00	6.0	5.0	5.0	16.0
M36	0.85	4.0	2.00	6.0	5.0	8.0	16.0
M41	1.10	4.25	2.00	6.75	3.75	5.0	14.25
M42	1.10	3.75	1.15	1.50	9.50	8.25	20.5
M43	1.20	3.75	1.60	2.75	8.00	8.25	18.75
M44	1.15	4.25	2.00	5.25	6.50	12.00	18.25
M45	1.25	4.25	1.60	8.25	5.0	5.50	18.25
M46	1.25	4.00	3.20	2.00	8.25	8.25	18.0

[a] Normal ranges of manganese, silicon, phosphorus, and sulfur are assumed (see Steel). The balance is Fe in all cases.

[b] $W_{eq} = 2\ (\%\ \mathrm{Mo}) + \%\ \mathrm{W}$.

[c] Widely available.

[d] Also contains 1.25% columbium.

be exercised during this treatment. In general, M-type tool steels are more popular, representing ca 85% of all tool steels, because they are less expensive by ca 30% than the corresponding T-type steels.

High speed steel tools are available in cast, wrought, and sintered forms. Improper processing of both cast and wrought products can lead to undesirable microstructure, carbide segregation, formation of large carbide particles, significant variation of carbide size, and nonuniform distribution of carbides in the matrix. Such a material is difficult to shape by grinding and causes wide fluctuations of properties, inconsistent performance, distortion, and cracking.

A processing technique introduced in the late 1960s involves atomization of the prealloyed molten tool steel alloy into fine powder, followed by consolidation under hot isostatic pressure (HIP) (20–23). This technique, termed consolidation by powder metallurgy (CPM), when combined with suitable hardening and tempering, provides a microstructure consisting of a uniform and fine dispersion of carbides in a fine-grained, tempered, martensite matrix. For example, a mean of 1.3 μm and a maximum of 3.5 μm for carbide grain size results from CPM compared to a mean of 6.2 μm and a maximum of 34 μm by conventional cast and wrought processes.

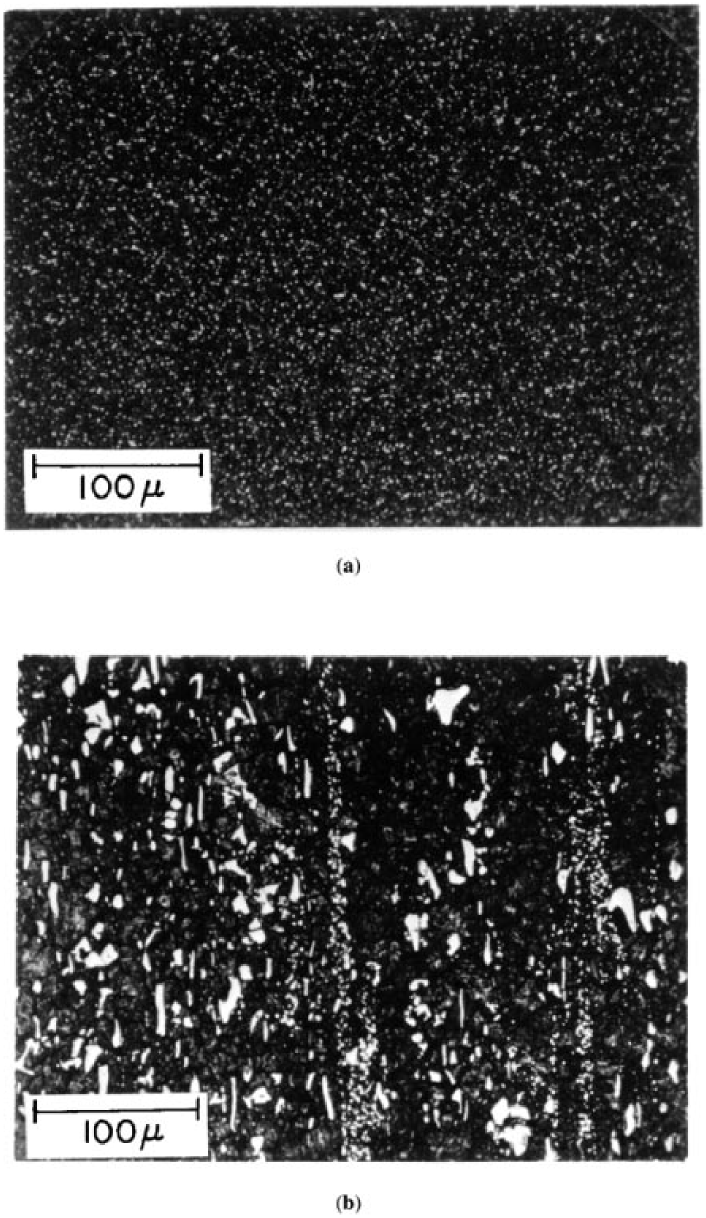

Fig. 4. Microstructure of AISI T15 tool steel (quenched and tempered) produced (**a**) from particles and (**b**) by the conventional technique (picral etch). In (**a**), the median and maximum carbide sizes are 1.3 and 3.5 mm, respectively; in (**b**), 6.2 and 34 mm, respectively.(Courtesy of Crucible Steel Company.)

Tool steels made in this manner grind more easily, especially the highly alloyed tool steels, with grinding ratios two to three times better; exhibit more uniform properties; and perform more consistently (24). Also, highly alloyed tool steels that can attain HRC 70 cannot be made by the conventional casting or hot forming processes but can be made by CPM. Because of the fine size of the carbides present in tool steels made by CPM, tools

Table 6. Chemical Compositions of Equivalent Grades With and Without Cobalt

	Chemical composition,[a] nominal %						
HSS Type	C	Cr	V	W	Mo	Co	W_{eq}[b]
T 15	1.55	4	5	12.25		5	12.25
Co-less T 15	1.08	4	5	12.5	6.5		25.5
M 42	1.1	3.75	1.1	1.5	9.5	8	20.5
Co-less M 42	1.3	3.75	2.0	6.25	10.5		27.25

[a] Balance is Fe in all cases.
[b] $W_{eq} = 2\ (\%\ Mo) + \%\ W$.

made of this material have significant edge strength and provide edge sharpness during cutting, such as in end milling. Consequently, material made by this process is extensively used to produce relatively inexpensive tools, such as drills, milling cutters, and taps, as well as expensive form tools such as broaches, shaper cutters for gears, and various dies for metal forming applications. Tool steels up to HRA of 70 can be obtained using high Co (up to 20%) and high vanadium carbide (VC) (also up to 20%). This would, however, be at the expense of significant loss of toughness. Tool steel technology has matured. Improvements in cleanliness of tool steels, ie, control of the composition of tramp elements, and tighter tolerances on the chemical composition, etc, are underway, as is improvement in the overall quality of the product.

The heat-treatment procedure generally consists of first preheating the HSS tool steel to 730–840°C, then heating rapidly to 1177–1220°C for 2–5 min to fully austenitize the steel, followed by quenching, initially in a suitable molten salt bath to a certain intermediate temperature (ca 600°C), and then cooling in air (16, 17, 19–23, 25). This treatment is followed by single or double tempering, where the steel is heated to 540–590°C for ca 1 h and then air-cooled to produce a tempered martensite structure containing unreacted larger carbides, and to relieve residual stresses.

Shortages and escalating costs of Co in the 1970s prompted tool-steel producers to seek an appropriate substitute. Hot hardness can be maintained without Co by appropriate increases of Mo–W or V content, or both (26). Higher concentrations of these latter elements in the matrix provide equivalent solid–solution strengthening at elevated temperatures. The compositions of steel grades with and without Co, yielding similar performance, are given in Table 6 (26). Micrographs of heat-treated (quenched and tempered) AISI M-42 tool steels with and without Co are shown in Figure 5. Despite heavy competition from cemented carbide, coated carbide, and ceramic tool materials, as of this writing (ca 1997) HSS accounts for the largest tonnage of tool materials used because of its unique properties (chiefly the toughness and the fracture resistance), flexibility in fabrication, and the fact that many cutting operations have to be conducted at a low enough speed range for HSS to perform efficiently and economically.

HSS tools are used mostly for low speed, heavy-duty applications. Thus built-up edge, adhesion of the chips to the tool, and high friction are the primary concerns for these tools rather than high tool temperatures. Consequently, when thin coatings on cemented carbides were developed in the late 1960s, the possible application of such coatings for HSS were considered. In fact, coatings for HSS tools preceded the use of coatings on cemented carbides (27, 28). The chemical vapor deposition (CVD) technique used for coatings on cemented carbide requires the tools to be heated to temperatures from 950–1050°C. Application of such temperatures would be unsuitable for HSS tools, altering the metallurgical structure and consequently the properties of the HSS substrate material. Thus a low (450°C) temperature technique, known as physical vapor deposition (PVD) is used to provide a thin (ca 5-μm) coating of TiN on HSS. This coating more or less provides both the needed protection against metal buildup on the tool and low frictional conditions. Consequently, TiN-coated HSS tools are gaining in popularity in large part because of the improvement in tool life from three to ten times when the tool is used in the speed range capability of HSS tools. The substrate material of TiN-coated HSS tools will be affected (softened) when the tool is operated at higher cutting speeds than those recommended. TiN coating

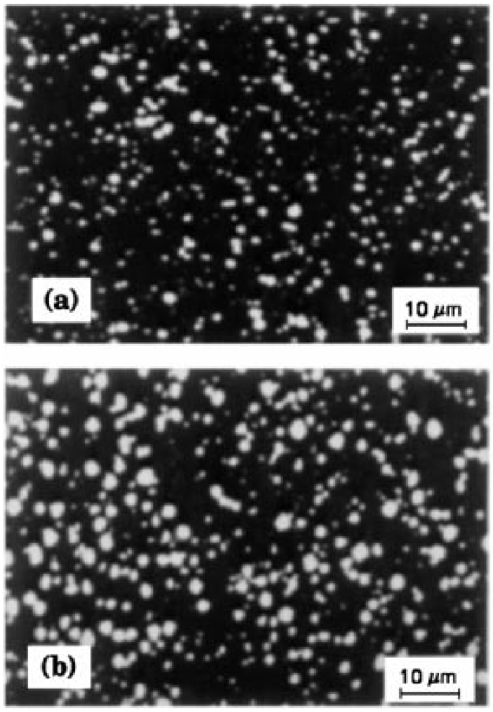

Fig. 5. Micrographs of the microstructure of fully hardened and tempered tool steels produced by the powder metallurgy technique, showing uniform distribution and fine carbide particles in the matrix. (**a**) M-42 (see Table 6) and (**b**) cobalt-free AISI T-15 having a higher concentration of fine carbide particles in the matrix.

also has a golden color that is aesthetically attractive and enables determination of the extent of tool wear. Thus TiN coatings are considered for wear applications where decorative value is important such as for watch cases. Uncoated HSS tools are used widely but upward of 40% of HSS tools are coated and this percentage is expected to increase.

A newer tool steel material having a fine grain size of TiC (40–55%) in a steel matrix (45–60%) with several unique characteristics was developed. Additional Cr (3–17.5%), Mo (0.5–4%), Ni (0.5–12%), Co (5–5.7%), Ti (0.5–0.7%), and C (0.4–0.85%) are made to provide solid solution strengthening as well as hot hardness of the matrix material (29–31). This material, which combines the hardness (consequently, the wear resistance) of cemented carbides with the heat treatability of HSS, responds to heat treatment, such as annealing and quench hardening, and can be machined in the annealed condition. The material is produced by initially compacting TiC powder in a steel die. The resulting porous compact is sintered at high temperature and subsequently infiltrated with molten steel under vacuum. The upper limit of TiC content is determined by the degree of machinability desired in the annealed condition. The microstructure of this material in the annealed condition shows well rounded carbide grains in a spheroidite steel matrix and in the quenched condition in a fine martensite matrix. The relatively wide separation between carbide particles in the annealed condition accounts for its good machinability.

Hardness in the annealed condition of the TiC in a steel matrix material is ca 69 HRA and after heat treatment ca 86.5 HRA. By compacting the TiC powder at higher pressure prior to infiltration, closer spacing and a harder microstructure result. Similarly, higher TiC content also leads to closer spacing. The transverse rupture strength (TRS) is ca 2068 MPa (300 ksi). The modulus of elasticity is 303 GPa (44×10^6 psi). TiC is also

Table 7. Composition and Properties of Some Representative Grades of Cemented-Carbide Tools[b][a]To convert J to ft·lbf, divide by 1.356.

	Composition, wt %				Properties			
Grade	WC	TiC	TaC	Co	Grain size	Density g/cm^3	HRA[c]	TRS[d], MPa[e]
				Nonsteel grades[g]				
roughing	94			6	coarse	15.0	91	2210
general purpose	94			6	medium	15.0	92	2000
finishing	97			3	fine		92.8	1790
				Steel grades[h]				
roughing	72	8	11.5	8.5	coarse	12.6	91.1	1720
general purpose	71	12.5	12	4.5	medium	12.0	92.4	690
finishing	64	25.5	4.5	6	medium	9.9	93.0	130

[b] Ref. 8.

[c] Rockwell hardness A scale.

[d] Transverse rupture strength.

[e] To convert MPa to psi, multiply by 145.

[f] To convert GPa to psi, multiply by 145,000.

[g] C-1 to C-4.

[h] C-5 to C-8.

Table 7. (Continued)

	Properties						
Grade	Elastic modulus *E*, GPa[f]	Impact strength, J[a]	Compressive strength, MPa[e]	Tensile strength, MPa[e]	Relative abrasion resistance, vol loss/cm^3	Thermal conductivity, W/(m·K)	Thermal expansion, per ($^\circ C^{-1} \times 10^{-6}$)
			Nonsteel grades[g]				
roughing	640	16	5170	1520	15	120	4.3
general purpose	650	16	5450	1950	35	100	4.5
finishing	610	12	5930	1790	60		4.3
			Steel grades[h]				
roughing	560	11	5170		8	50	5.8
general purpose	570	9	5790		7	35	5.2
finishing	460	5	4900	480	5		5.9

chemically more stable when machining steels at medium speeds. In the annealed condition this material is soft enough that it can be machined to shape. After heat treatment, such as austenitizing followed by oil quenching and tempering, tempered martensite structure is formed in the binder phase, resulting in a significant increase in the hardness of this material. The additional feature of this material is that owing to comparable thermal expansion coefficients, it can be easily brazed or welded to the steel substrate without any danger of cracking.

Another HSS tool material, similar to the TiC in a steel matrix, is comprised of 30–60% of submicrometer (ca 0.1 μm) TiN hard phase dispersed in a heat treatable steel (Coronite) (32). It can be seen that the percentage of hard phase, TiN in this alloy is higher than in HSS but less than the lowest limit of cemented carbide. It is thus harder than any conventional HSS, but tougher than most cemented carbides. At the same time, the fine grain size of TiN ensures excellent edge strength especially for milling cutters, drills, etc used in the machining of steels. Because TiN is also chemically more stable when machining steels, this combined material should fill the gap between HSS and cemented carbide. This material can be heat-treated and ground more easily than the cemented TiC counterpart. Tools consist of a steel (HSS or spring steel) core on which the TiN–HSS material is pressed using powder metallurgy technology to comprise about 15% of the diameter. The outer surface can then be coated with TiCN or TiN by PVD.

2.3. Cast-Cobalt Alloys

Cast-cobalt alloys were introduced about the same time as HSS for cutting tool applications. Popularly known as Stellite tools, these materials are Co-rich Cr–W–C cast alloys having properties and applications in the intermediate range between HSS and cemented carbides. Although comparable in room-temperature hardness to HSS tools, cast-cobalt alloy tools retain their hardness to a much higher temperature (see Fig. 2) and hence can be used at higher (25%) cutting speeds than HSS tools. Cast-cobalt alloys contain a primary phase of Co-rich solid solution (instead of Fe in HSS) strengthened by Cr and W, and dispersion-hardened by complex, hard, refractory carbides of W and Cr (33, 34). Unlike HSS, cast-cobalt alloys are hard as cast, and cannot be softened or hardened by heat treatment. Cast-cobalt alloys have, however, been phased out owing to the high cost of Co, safety in handling Co-base alloys, and availability difficulties.

2.4. Cemented Carbides

Tungsten carbide was first synthesized in the 1890s, but satisfactory methods for fabricating this material in bulk form with adequate strength in the form of cutting tools, dies, or wear parts were not developed until many years later (35). The main impetus for such a development was to replace the expensive diamond dies used in the manufacture of tungsten wire for lighting. The first cemented-carbide tool material, introduced in Germany in the mid-1920s, was an unalloyed tungsten carbide, WC, in a Co binder (36, 37). The material was called Widia for *wie diamente*, like a diamond, and introduced at the Leipzig Trade Fair in 1927 (38). There are some 200 cemented carbide manufacturers worldwide as of the mid-1990s. Prominent among them are Sandvik, Kennametal, Velerite, Krupp Widia, Carboloy, Carmet, Teledyne Firth Sterling, Stellram, Mitsubishi, Sumitomo, Toshiba Tungaloy, and Iscar (12).

Cemented carbides are a class of tool material containing a large-volume fraction ($\geq 90\%$) of fine-grain, refractory carbides (WC or solid solutions of carbides of W, Ti, and Ta or Nb, and TiC) in a metal binder (Co for the first two types and Ni–Mo for TiC) produced by cold pressing followed by liquid-phase sintering (39–44) (see Carbides, industrial hard carbides). The binder material is chosen so that it wets the carbide to form a good bond thus enabling the carbide to be sintered into a dense mass. This introduces a ductile component into the microstructure, thereby increasing the material toughness. Moreover, by varying the amount of the binder phase, cemented carbides tool materials of different toughness values can be obtained. For metal-forming dies and wear parts the percentage of the binder can be quite high, as much as 25%. In cutting tool applications, the binder is also expected to provide refractoriness and chemical stability at high temperatures. Some tungsten is deliberately allowed to alloy with Co to provide these features, in addition to the solid solution strengthening. Where other properties, such as higher conductivity, or more chemical stability, or higher strength are required, suitable alloying elements can be added to the binder phase (see Carbides, cemented carbides; Refractories).

The function of Co in cemented WC is to act as a medium in which carbide grains can grow together to form a skeletal structure (39), not merely act as a binder. The carbide particles tend to grow together and may actually bond at several locations in the grain, forming a skeletal structure. Thus the continuous phase in cemented WC or multicarbide is the carbide phase which accounts for the high modulus, which is significantly higher (ca three times) than that of HSS.

Cemented carbides differ from HSS in many important respects. They are much harder, chemically more stable, and superior in hot hardness. They are also generally lower in toughness than HSS. They can be used at cutting speeds three to six times higher than HSS. Carbide is the continuous phase in cemented carbides just as the metallic phase (FC) is in HSS. As a result, the Young's modulus, E, of cemented carbide is two to three times that of HSS (414–689 GPa $(60 - 100 \times 10^6$ psi)). Consequently, cemented carbide is two to three times stiffer than HSS. Furthermore, a specific grade of cemented carbide can be used to machine a specific work material, thus minimizing chemical interaction between the tool and the work material. This is possible in cemented carbides because the chemistry of the primary (carbide) phase can be altered to provide the needed stability. Cemented carbides have a lesser tendency for adhesion, except at low speed and heavy loads, but are more brittle and expensive to fabricate and shape than HSS. A wide range of hard refractory coatings (qv) can be deposited with some reduction in TRS of the substrate in the CVD process. The strategic metals W, Co, and Ta are used extensively in cemented carbides.

Most cemented-carbide tools are WC-based and have Co as the binder. Other carbide tool materials based on TiC having a Ni–Mo binder were developed primarily for high $(>300 - 500$ m/min$)$ speed finish machining of steels and gray cast irons for automotive applications.

In the machining of cast iron, cemented straight tungsten carbide tool material exhibits a long tool life even at three to six times the cutting speeds used with HSS. When machining steels, however, cemented WC develops a deep crater on the tool face owing to chemical interactions, thus leading to rapid wear. Improved stability of solid solutions of multicarbides of W–Ti or W–Ti–Ta over a mechanical mixture of WC, TiC, and TaC or unalloyed WC in providing considerable resistance to crater wear when machining steels has been observed (45, 46). A unique process for the production of solid solution carbides of two or more refractory carbides has been developed (47–51). Different grades of cemented carbides were obtained by varying the Co content, the amount of different carbides, and the carbide grain size (see Table 7). The higher Co grades or coarser carbide size grades are tougher but less hard; the more complex carbides are harder, and chemically more resistant (especially to steels), but weaker than WC–Co alloys. Production of cemented tungsten carbide is a mature technology. Good control on quality is maintained irrespective of the tool manufacturer.

Figure 6 contains micrographs of representative nonsteel machining grades of cemented carbides (roughing, general-purpose, and finish-machining grades, respectively) containing unalloyed WC with decreasing grain size or Co content. Figure 7 contains micrographs of similar grades for steels containing different amounts of complex multicarbides in a Co binder. Progressing from a roughing to a finishing grade, the hardness increases, toughness decreases, and resistance to high temperature deformation and wear resistance increases. The variation of hardness, transverse rupture strength, impact strength, and elastic modulus with percent Co binder content for straight cemented-WC grades are shown in Figure 8 (52). Hardness and elastic modulus decrease with an increase in Co content; the impact strength and transverse rupture strength increase.

Submicrometer grain size $(<1-\mu$m and typically in the range of 0.1–0.5-μm) cemented tungsten carbide tool materials were developed in the late 1960s to increase toughness and edge strength (53, 54). Fine dispersions of small amounts (0.5%) of submicrometer chromium carbide restrict grain growth of WC in this alloy. Because of the fine grain size the grain boundary area of this material is significantly higher than a similar material having larger grain size. Consequently, higher binder content can be used without sacrificing hardness, but increasing the toughness. The transverse rupture strength (TRS) of the submicrometer cemented tungsten carbide can be up to 2757 MPa (400 ksi) which is close to that of HSS. But, the hardness of this carbide is significantly higher, 91.5 R_A, compared to 70 for HSS (53, 54).

Sometimes cemented carbide tools are used not only for hardness and wear resistance but also for high modulus or stiffness. For example, in end mills used in high speed machining of aluminum alloys, the deflection of the tool can affect the performance of the tool considerably. This includes chatter or vibrations of the tool, tolerance, and finish requirements. In such circumstances, a solid carbide provides nearly three times the stiffness of HSS as well as providing the wear resistance required at the cutting edge, thus overcoming some of the problems experienced in the shop floor and at the same time increasing the productivity significantly. A similar situation involves long-boring bars used in steel matching.

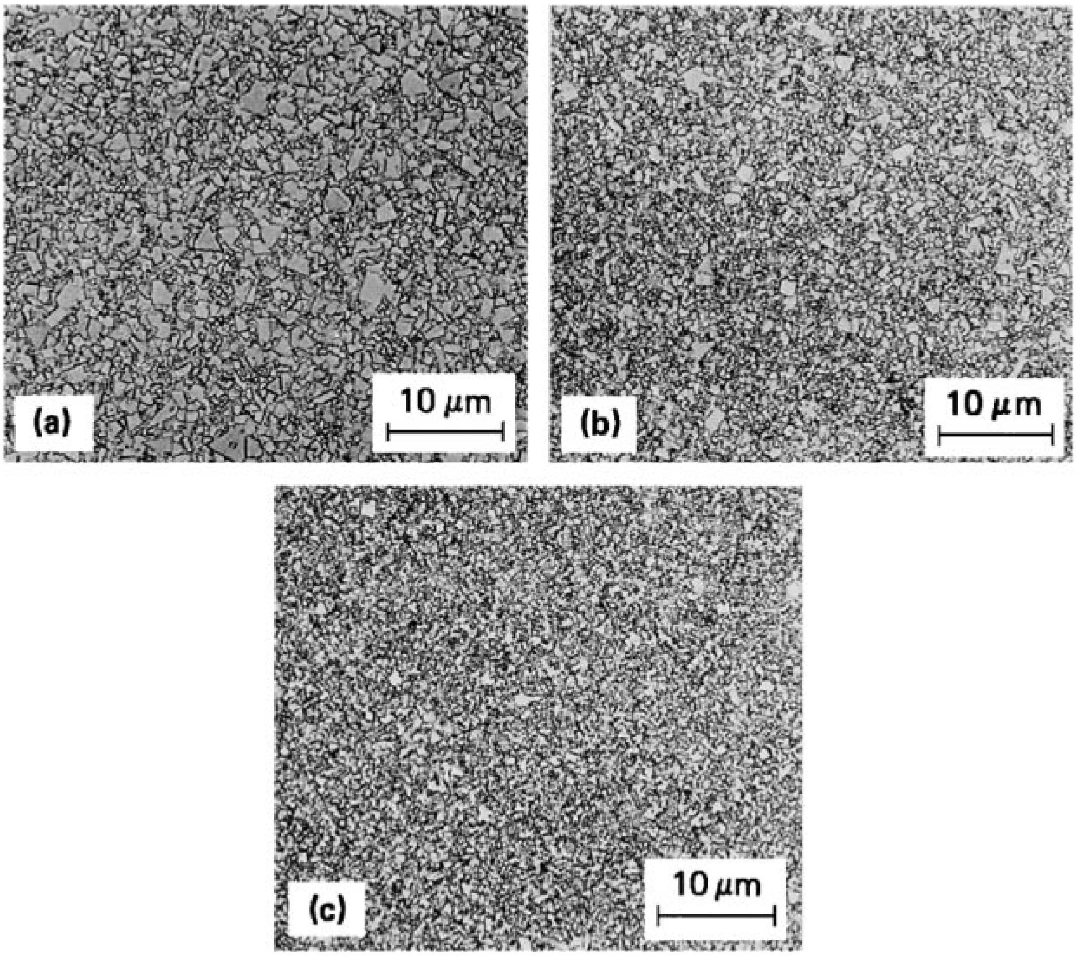

Fig. 6. Micrographs of three representative grades of cemented, unalloyed WC in a Co binder for (**a**) roughing, (**b**) general purpose, and (**c**) finish-machining of materials other than steels (9).

There are at least four different classification systems for cemented carbides (7, 12). The U.S. system is based on relative performance; the U.K. system is based on properties, and the former USSR system on composition; the fourth system, widely used in Europe and supported by the ISO, is based on application and chip form. In this article, the U.S. system and the ISO system are briefly reviewed.

In the United States, the C-classification (C-1 to C-8) for cemented carbide tools, used unofficially for machining applications, was originally developed by the automobile industry to obtain a relative performance index of tools made by different tool producers. This is by far the simplest system. The grades are broadly divided into two classes (C-1 to C-4 and C-5 to C-8), according to the type of work material to be machined. Grades C-1 to C-4 are recommended for machining nonsteels, ie, cast iron, nonferrous alloys, and nonmetallics (nonsteel workmaterials), whereas C-5 to C-8 are recommended for machining carbon steels and alloy steels. Although the grades to be used for machining other difficult-to-machine materials, eg, the titanium alloys and Ni-base and Co-base superalloys, have not been specified explicitly in this classification, the nonsteel grades C-1 to C-4 are applicable for machining nonferrous alloys. Many users of this classification system are not

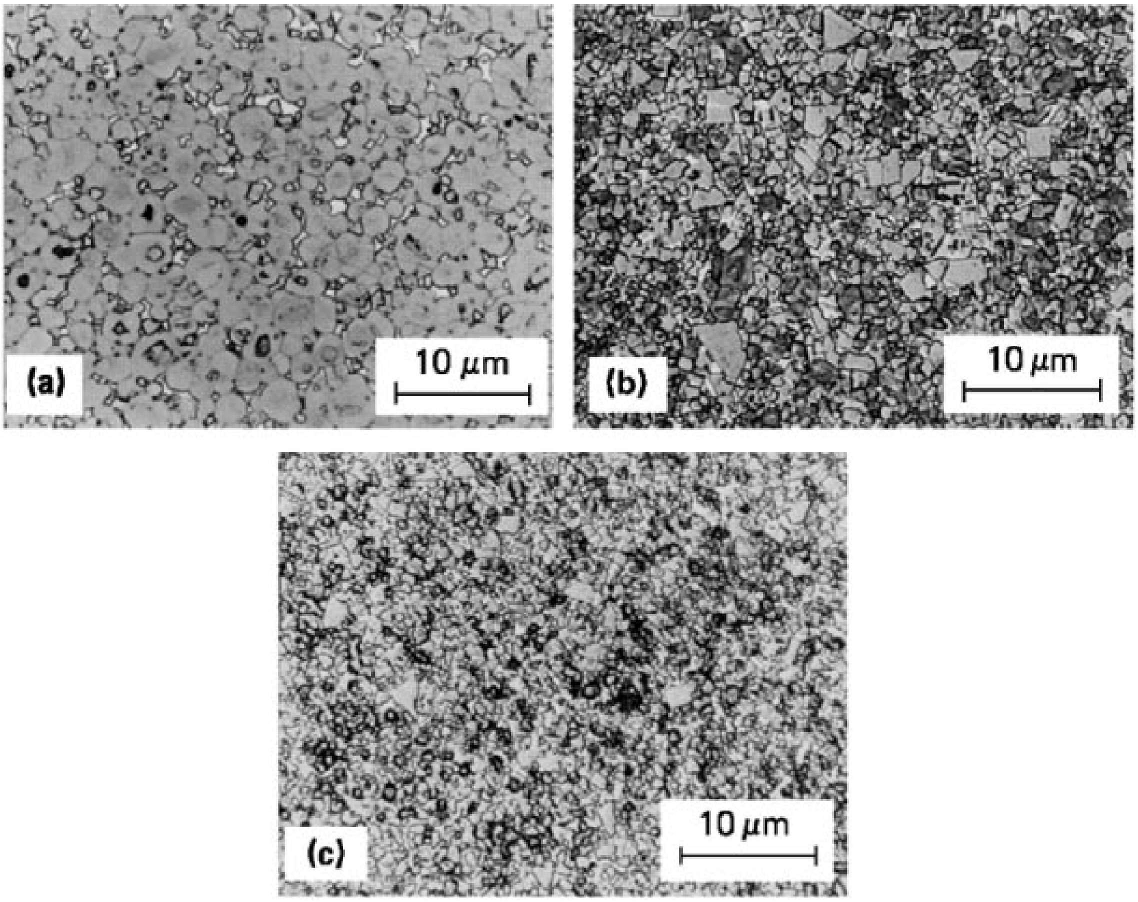

Fig. 7. Micrographs of representative grades of cemented carbides for steels containing different amounts of solid-solution multicarbides in a Co binder (9). (**a**) Roughing, (**b**) general purpose, and (**c**) finish-machining grades.

familiar with this, as they consider grades C-1 to C-4 as cast-iron grades, and not as grades for machining materials other than steels. In general, the nonsteel grades are straight WC in a Co binder, whereas the steel grades are solid solutions of multicarbides in a Co binder.

Within each class, ie, C-1 to C-4 and C-5 to C-8, each grade is distinguished by the type of machining operation: C-1 and C-5 for roughing, C-2 and C-6 for general purpose, C-3 and C-7 for semifinishing, and C-4 and C-8 for precision-finishing operations. In general, from grades C-1 to C-4 or C-5 to C-8 within each class, the shock resistance decreases, hardness increases, high temperature deformation resistance and wear resistance increase, and the Co content and carbide grain size decrease. Roughing and general-purpose grades require more toughness to withstand heavy loads, whereas finishing and semifinishing grades require a high temperature deformation-resistant and a wear-resistant sharp edge. At one time, each tool producer associated one or more carbide grades with the eight grades in the C-classification. However, this comparison involves competition only of grades identified within each class, eg, C-1 or C-5 by different manufacturers, and hence there is a trend to disassociate from this classification. Individual cemented-carbide producers deviate from this rule slightly (in the carbide grain size and Co content) to gain a competitive edge.

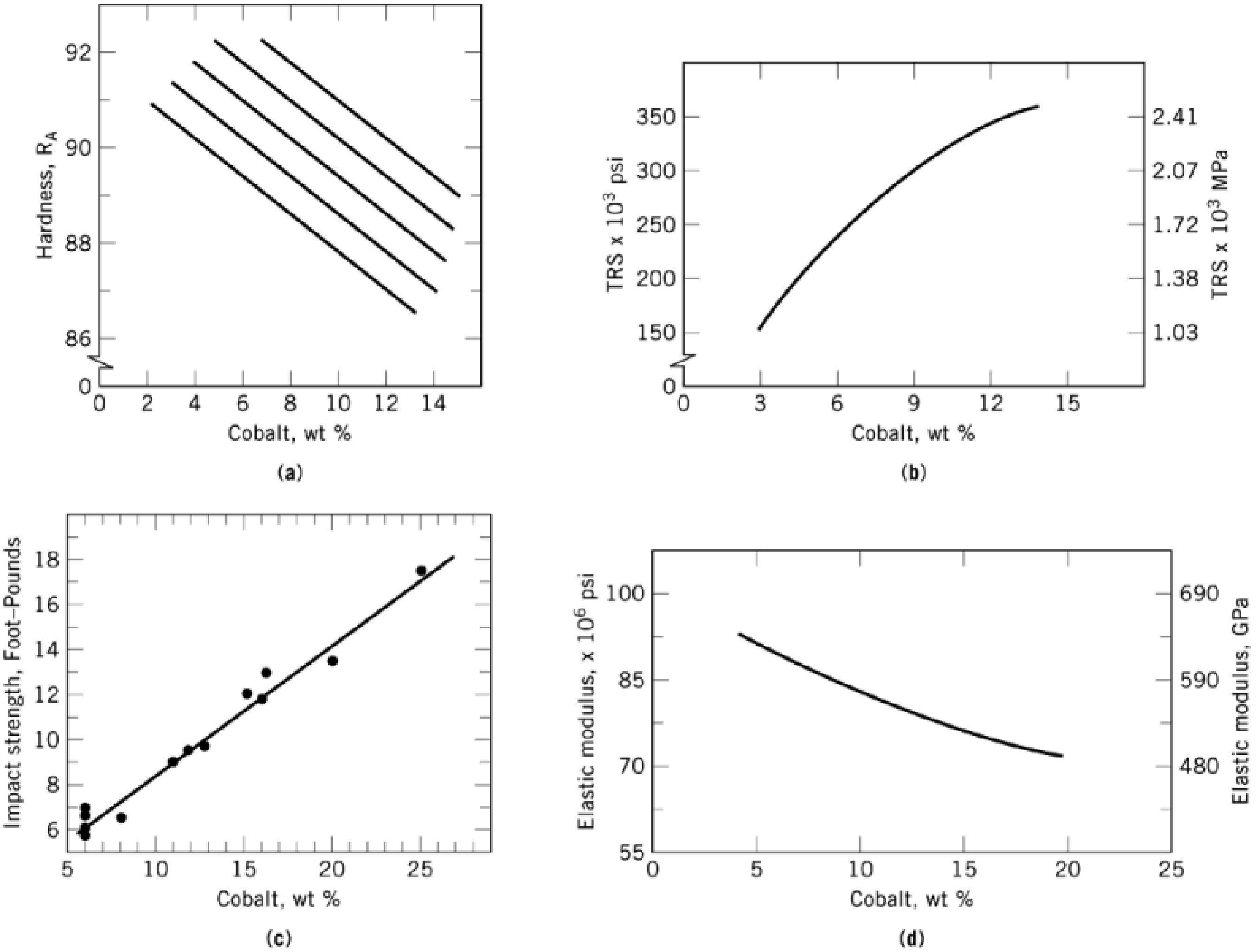

Fig. 8. Variation with percent Co binder content for cemented WC grades of (**a**) hardness, where the numbers represent carbide grain size in μm; (**b**) transverse rupture strength (TRS); (**c**) impact strength; and (**d**) elastic modulus (52).

The ISO classification for cemented-carbide cutting tools is given in Table 8 (12). Some prior knowledge of machining is expected in consulting this table, which is broadly divided into three categories and, for convenience as well as easy identification, color coded when used on the shop floor. P-grades (blue) are highly alloyed multicarbides used mainly for machining hard steels and steel castings; M-grades (yellow) are low alloy multicarbide alloys which are multipurpose nonsteel grades used for machining high temperature alloys, low strength steels, gray cast iron, free machining steels, and nonferrous metals and their alloys; and K-grades (red) are straight WC grades for machining very hard gray cast iron, chilled castings, nonferrous metals and their alloys, and nonmetallics such as plastics, glass, glass–epoxy composites, hard rubber, and cardboard. Details of the work material to be machined, the type of cutting operation, eg, continuous vs intermittent or roughing vs finishing, and the type of chip formed are included in this classification. There are six basic categories in the P-type, four in the M-type, and five in the K-type. Carbide grades differing from these basic categories can be designated by appropriate in-between numbers within each class, eg, between P 01 and P 10. Thus, the coated grades that do not fall under the basic categories can be placed in these categories. The lower numbers, eg, P 01, M 10, or K 10, are for higher speed, finishing (lighter cut) applications (harder with low Co–finer carbide grain size), and the higher numbers (P 50, M 40, or K 40) are for lower speed, roughing (heavier cut) applications (tougher with higher Co–coarser carbide grain size).

The original objective of the ISO classification was to issue detailed standards for cemented carbides in terms of microstructure, composition, and properties for quality control and performance reliability. This objective, however, is yet to be realized. Increased emphasis on worldwide implementation of ISO 9000 standards and globalization of manufacturing, may lead the industry-at-large to adopt the ISO classification.

In selecting a carbide grade for a given application, the following general guidelines should be followed: the grade with the lowest Co content and the finest grain size consistent with adequate strength to eliminate chipping should be chosen; straight WC grades can be employed if cratering, seizure, or galling is not experienced and for work materials other than steels; to reduce cratering and abrasive wear when machining steels, TiC grades are preferred; and for heavy cuts in steel where high temperature and high pressure deform the cutting edge plastically, a multicarbide grade containing W–Ti–Ta(Nb) with low binder content should be used.

Table 8. ISO Classification of Cemented Carbide Tools According to Use[a]

Main groups of chip removal		Groups of application			Increase or decrease in characteristic	
Symbol (color)	Categories of material to be machined	Designation	Material to be machined[b]	Use and working conditions	Of cut	Of carbide
P (blue)	ferrous metals with long chips	P 01	steel, steel castings	finish turning and boring; high cutting speeds; small chip section; accuracy of dimensions and fine finish; vibration-free operation	↑ Increasing speed Decreasing feed	Wear resistance Decreasing toughness
		P 10	steel, steel castings	turning, copying, threading, milling; high cutting speeds; small or medium chip sections		
		P 20	steel, steel castings, malleable cast iron with long chips	turning, copying, milling; medium cutting speeds; medium chip sections; planing with small chip sections		
		P 30	steel, steel castings, malleable cast iron with long chips	turning, milling, planing; medium or low cutting speeds; medium or large chip sections; machining in unfavor-able conditions[b]		
		P 40	steel, steel castings with sand inclusion and cavities	turning, planing, slotting; low cutting speeds; large chip sections; possibility of large cutting angles for machining in unfavorable conditions[b] and work on automatic machines		
		P 50	steel, steel castings of medium or low tensile strength, with sand inclusion and cavities	for operations demanding very tough carbides; turning, planing slotting; low cutting speeds; large chip sections; possibility of large cutting angles for machining in unfavorable con-ditions[b] and work on automatic machines		

Table 8. ***Continued***

Main groups of chip removal		Groups of application			Increase or decrease in characteristic	
Symbol (color)	Categories of material to be machined	Designa-tion	Material to be machined[b]	Use and working conditions	Of cut	Of carbide
M (yellow)	ferrous metals with long or short chips and nonferrous metals	M 10	steel, steel castings, manganese steel, gray cast iron, alloy cast iron	turning; medium or high cutting speeds; small or medium chip sections	↑ Increasing speed Decreasing feed	Wear resistance Decreasing toughness
		M 20	steel, steel castings, austenitic or manganese steel, gray cast iron	turning, milling; medium cutting speeds; medium chip sections		
		M 30	steel, steel castings, austenitic steel, gray cast iron, high temperature resistant alloys	turning, milling, planing; medium cutting speeds; medium or large chip sections		
		M 40	mild free cutting steel, low tensile steel, non-ferrous metals and light alloys	turning, parting off, particularly on automatic machines		
K (red)	ferrous metals with short chips, nonferrous metals, and non-metallic materials	K 01	very hard gray cast iron, chilled castings of over 85 Shore, high silicon–aluminum alloys, hardened steel, highly abrasive plastics, hard cardboard, ceramics	turning, finish turning, boring, milling, scraping		
		K 10	gray cast iron over 220 Brinell, malleable cast iron with short chips, hardened steel, silicon aluminum alloys, copper alloys, plastics, glass, hard rubber, hard card-board, porcelain, stone	turning, milling, drilling, boring, broaching, scraping	↑ Increasing speed Decreasing feed	Wear resistance Decreasing toughness
		K 20	gray cast iron up to 220 Brinell, nonferrous metals (copper, brass, aluminum)	turning, milling, planing, boring, broaching, demanding very tough carbide		
		K 30	low hardness gray cast iron, low tensile steel, compressed wood	turning, milling, planing, slotting, for machining in unfavorable conditions[b] and with the possibility of large cutting angles		
		K 40	soft wood or hard wood, nonferrous metals	turning, milling, planing, slotting, for machining in unfavorable conditions[b] and with the possibility of large cutting angles		

[a] Ref. 12.

[b] Raw material or components in shapes which are awkward to machine: casting or forging skins, variable hardness, variable depth of cut, interrupted cut, work subject to vibrations, etc.

Composition, microstructure, and performance of cemented carbides depend on Co binder content, carbide grain size, and type and composition of various carbides. With increasing Co content, toughness as reflected by the transverse rupture strength (TRS) and impact strength increase, whereas hardness, Young's modulus, and thermal conductivity decrease (see Fig. 8) (52). Finer grain size gives a harder product than coarser grain-size carbides. Multicarbides increase chemical stability and both room-temperature and hot hardness, whereas TiC addition to WC controls crater wear, especially for machining steels. The proper grade of cemented carbide for a given work material should provide adequate crater-wear resistance, abrasion resistance, and toughness to prevent microchipping of the cutting edge.

Another type of carbide tool material, developed by the Ford Motor Company in the late 1950s for high speed ($>300-400$ m/min) finish machining (low feed) of steels, is based on cemented TiC in a Ni–Mo binder (55–59). According to the ASTM definition, cermet (*cer*amic–*met*al) is an acronym to designate a heterogeneous combination of metal(s) or alloy(s) with one or more ceramic phases in which the latter constitutes approximately 15–85% by volume and in which there is relatively little solubility between the metallic and ceramic phases at the preparation temperature. Whereas cemented tungsten carbide also comes under this category, the common practice is to consider only TiC materials as cermets. Actually, TiC-based cermets with a Ni binder were explored in Germany in the 1930s (40, 45, 46). Because of poor wetting of TiC by Ni, this material was not as strong as cemented WC and hence not as effective as a cutting tool material.

A breakthrough occurred in the development of cermets when in 1956 (60) additions of Mo to TiC–Ni cermet were shown to improve wetting of the carbide by forming a mixed carbide shell (Ti,Mo)C around the TiC grains, thereby inhibiting carbide coalescence and grain growth. The resulting microstructure gives improved hardness and impact resistance. However, this material was not as hard and tough as the cemented multicarbide counterpart. Because of the wide popularity enjoyed by cemented carbide tools and the initial rather negative reputation gained by cemented TiC for being relatively brittle and easy to chip or fracture, there was a reluctance to change from the cemented WC tools to cemented TiC tools.

Further advancement of this cermet is based on the following: (*1*) improvements through the additions of other carbides, such as MoC, TaC, and WC as well as Co binder; (*2*) significant additions of TiN to TiC either as separate phases or as titanium carbonitride, TiCN, resulted in TiC–TiN cermets being widely used commercially, especially in Japan; (*3*) modification of the composition of the cermet by adding Al to the alloy, which precipitates fine Ni_3Al particles in the binder phase for improving the elevated temperature strength, similar to the strengthening effect found in nickel-base superalloys; and (*4*) TiN coating, preferably by physical vapor deposition (PVD), on the cermet tools. With these additions the hot hardness, transverse rupture strength (TRS), oxidation resistance, and thermal conductivity are significantly increased. Table 9 shows a comparison between the original TiC–Ni–Mo cermet and the cermet containing TiN, WC, TaC, and Co (61). The higher TRS provides better edge strength and chipping resistance, whereas higher thermal conductivity provides thermal shock resistance, both of which have limited the application of this material for a long time.

Table 9. A Comparison of Properties of Cermets

Composition of cermet	HV,[a] at 1000°C	TRS, 900°C, N/mm^2	Oxidation resistance wt gain after 1 h at 1000°C, mg/cm^2	Thermal conductivity at 1000°C, W/(K·m)
TiC–16.5% Ni–9% Mo	500	1050	11.8	24.7
TiC–20% TiN–15% WC–10% TaC–5.5% Ni–11% Co–9% Mo	650	1360	1.66	42.3

[a] HV = Vicker's hardness.

Compositions and properties of C-5 to C-8 steel grades of cemented TiC are given in Table 10. The TRS of cemented TiC is higher than that of most ceramics, but lower than that of cemented WC. Furthermore, the Young's modulus of cemented TiC, although double that of HSS, is ca 25% less than that of cemented WC. Like the cemented WC tools, cemented TiC is processed by cold pressing followed by liquid-phase vacuum sintering. Finishing steel-grade cemented TiC was originally used for high speed finish turning and boring of steels. The improved grades are used in the high speed milling of steels and malleable cast irons. Coatings on cemented tungsten carbide are used extensively; those on cemented titanium carbide have begun to be used more recently (62). Refractory, hard fiber-reinforced, and multilayer-coated cemented TiC are expected to be available for a range of steel and cast iron machining applications in the future.

Table 10. Composition and Properties of Steel Grades of Cemented Titanium Carbide[a]

	Composition, wt %			Properties			
Grade[d]	TiC	Ni	Mo	HRA[e]	TRS, MPa[b]	Young's modulus, GPa[c]	Density g/cm^3
roughing	67–69	22	9–11	91	1900	413	5.8
general purpose	72–74	17	9–11	92	1620	431	5.6
finishing	77–79	12	9–11	92.8	1380	440	5.5

[a] Ref. 9.
[b] To convert MPa to psi, multiply by 145.
[c] To convert GPa to psi, multiply by 145,000.
[d] C-5 to C-8.
[e] Rockwell hardness A scale.

The TiC–TiN cermets or titanium carbonitride cermets, especially those having coatings, cover the machining range between cemented carbides at one end and ceramics on the other. These have attained a significant degree of prominence in Japan, accounting for some 30% of the cutting tool market, mostly for high speed finishing (light cuts) of near net-shaped parts (63). This success is mainly the result of their high hot hardness (and consequently high deformation resistance), high wear resistance (both flank and crater wear), high resistance to metal buildup, and higher thermal conductivity and thermal shock resistance, compared to cemented TiC (64). Consequently, improved surface finish on the part, consistent part tolerances, high speed capability, and longer tool life result. Different grades of TiC–TiN were developed to cover roughing (ca 12% binder), general purpose (ca 10% binder), and finishing (ca 9% binder). In addition, coated grades having high binder content (ca 18% binder) were developed for interrupted cutting, such as milling. In Europe, the use of TiC–TiN cermets is growing steadily (ca 5%). In the U.S., however, cermets are a niche market as in the auto industry where productivity gains, superior part finish, and tolerance are critical and specialized high-speed, high-power precision machine tools are available.

Although cemented-carbide tools can be brazed, most of the carbide tools are available in insert form such as squares, triangles, diamonds, and rounds. These can be easily clamped on to the tool shank, thereby avoiding the problems and complexities associated with brazing. It is this feature that widely extended the applications of cemented carbide tools. Many cemented carbide grades are relatively less strong compared to HSS, especially when machining high strength materials. Thus these grades are used in such a manner that a square insert cuts with all its eight corners successively.

Chip breaking and disposal during cutting was not as serious a problem when manually operated machine tools were used. The clamped chip breakers, which were adequate to control the chip flow, are not acceptable for numerically controlled (NC) machine tools. Thus built-in (mould-in) chip grooves of various shapes and complexities have been developed for various applications, such as different work materials, different machining processes, and for roughing, semifinishing, and finishing. A myriad of chip groove geometries are available. These built-in chip breakers not only break chips but also facilitate chip disposal during machining. They also reduce the forces involved in machining, thereby improving the efficiency of cutting.

Cemented carbides are not generally recommended for low speed cutting operations because the chips tend to weld to the tool face and cause microchipping and there is no economic incentive to use them at lower speeds. However, for applications requiring higher stiffness, and higher wear resistance, such as broaches and shaper cutters, they are used extensively at lower speeds. Thin coatings of TiN made this application even more attractive. Cemented carbides are especially effective at higher speeds, generally in the 45–180 m/min range. This speed can be much higher (>300 m/min) for materials that are easier to machine, eg, Al alloys, and much lower (ca 30 m/min) for materials more difficult to machine, eg, Ti alloys. In interrupted cutting applications, edge chipping is prevented by appropriate choice of cutter geometry and cutter position with respect to the workpiece in such a way as to transfer the point of application of the load away from the tool tip. Finer grain size and higher Co content improve toughness in straight WC–Co grades and are considered desirable in materials used for interrupted cutting. Because of the high hardness of cemented carbides, they can be finished only by diamond grinding. Abusive grinding can lead to thermal cracks and poor performance.

To conserve the strategic materials (W, Co, and Ta) and reduce costs, recycling of used cemented-carbide inserts (so-called disposable or throwaway inserts) is growing steadily (65). Cobalt can be removed either by chemical leaching or by heating to high temperature (ca 1700–1800°C) in a vacuum to vaporize some of the Co and embrittle the rest of the material, leaving the carbide particles intact. The mass is then pulverized and screened to produce a fine powder. Other separation techniques include the zinc reclaim process commercialized in the 1970s (43). Alternatively, the cemented carbide inserts can be reground for applications where the actual size of the insert is not of critical concern. Several commercial fabricators provide regrinding services on a regular basis. In the extreme case, to conserve these materials economically, new techniques could be developed wherein the cemented carbide is used only at and near the cutting tips (65).

3. Coated Tools

The difficulty of machining many advanced materials presented challenges to the cutting tool industry, leading to the introduction of coated cemented carbides in the late 1960s (66–71) and coated HSS in the late 1970s. The technique commonly used for coated cemented carbides, chemical vapor deposition (CVD), requires that the tools be heated to ca 950–1050°C. At these temperatures, the metallurgical structure of HSS can be altered significantly. Thus, only coatings requiring substrate heating below the HSS transformation temperature (ca 450–500°C) can be applied for HSS. This is accomplished by a thin coating of TiN on HSS using physical vapor deposition (PVD) processes which are operated at lower temperatures (ca 400–450°C). Because most HSS tools are used for low speed applications, material buildup and friction on the tool face are the main considerations. TiN coating provides an acceptable solution to these problems because of its significantly higher hardness than HSS. Although rapid advances in coated cemented-carbide technology have taken place, coating technology for HSS is still limited to coating of TiN by PVD.

An analysis of the cutting process indicates that the material requirements at or near the surface of the tool are different from those of the tool body. The clearance surface, to be abrasion-resistant, has to be hard, and the rake face, to prevent chemical interaction, has to be chemically inert. The weakest link in the case of cemented carbide is the Co binder which is a soft metal and has a lower melting temperature than the carbide. A thin, chemically stable, hard, refractory binderless coating often satisfies these requirements. If the coating is too thick, it exhibits its bulk qualities, principally the brittleness. The tool body, by contrast, should have adequate deformation resistance to withstand high temperature plastic deformation of the nose and the body of the cutting tool under the conditions of cutting. These requirements are somewhat conflicting. Coated tool design is thus considered in terms of engineered materials. This methodology is also termed surface engineering. A thin (ca 5-μm) coating of TiC was developed for cemented-carbide tools in the mid-1960s. Patents were issued in the early 1970s (72–82). Over 200 U.S. patents have been issued to various tool manufacturers worldwide on the development of coatings on cutting tools as of the mid-1990s.

An effective coating should be hard; refractory; chemically stable; chemically inert to shield the constituents of the tool and the work-material from interacting chemically under the conditions of cutting; binder free; of fine grain size with no porosity; metallurgically bonded to the substrate with a graded interface to match the properties of the coating and the substrate; thick enough to prolong tool life but thin enough to prevent brittleness; free of the tendency of metal chips to adhere to or seize to the tool face; able to provide residual compressive stress; easy to deposit in bulk quantities; and inexpensive. In addition, coatings should have low friction and exhibit no detrimental effects on the substrate or bulk properties of the tool.

In order for the coating to adhere strongly to the substrate, several factors should be considered (83). These include mechanical, physical, and chemical compatibilities between the coating and the substrate. Because the tools are subjected to a high intensity of loading during cutting, the substrate must have adequate hardness and deformation resistance to support the coating without deformation. Otherwise, the coating becomes delaminated from the substrate owing to the development of interfacial tensile stresses. The stress level is intensified in the coating where the coating is stiffer than the substrate. These stresses increase with increasing load and with increasing mismatch in the modulus of elasticity. The relative in-plane normal stress levels in the coating and the substrate are proportional to the ratio of the elastic moduli. Thermal expansion mismatch between the coating and the substrate is another factor responsible for the incompatibility. Tensile stresses in the coating are most damaging. The stress becomes more tensile with increasing temperature when the thermal expansion of the substrate is greater than the coating. Alternatively, if the thermal expansion coefficient of the substrate is lower, then residual compressive stresses are induced which are beneficial to the adhesion of the coating to the substrate.

To avoid failure of the coating at the interface, the strength of the interface should be very high. Where the properties of the coating and the substrate are significantly different, as in the case of diamond and carbide substrate or alumina and carbide substrate, it is preferable to develop a graded interface to take into account factors promoting strong bonding. If chemical bonding is not feasible, then mechanical bonding involving interlocking of asperities should be considered prior to coating deposition. This may be accomplished by mechanical action or chemical etching. To improve adhesion, many manufacturers develop a graded interface between the coating and the substrate. For example, a graded interface with a carbon-rich layer of TiC or TiCN adjacent to the substrate followed by a series of varying layers of coating material wherein the content of carbon progressively decreases and nitrogen progressively increases with the surface layer rich in nitrogen content, TiN, has been developed (84).

Several refractory coatings (qv) have been developed including single coatings of TiC, TiN, Al_2O_3, HfN, or HfC, and multiple coatings of Al_2O_3 or TiN on top of TiC, generally deposited by CVD. TiC is used as a hard wear-resistant coating at low speeds. Multiple coatings prolong tool life, as the thickness of the coating can be increased from ca 5 to 10 μm without inducing brittleness, provide a strong metallurgical bond between the coating and the substrate by choosing appropriate coatings that would provide graded interface(s), and provide protection for machining a range of work materials. Figure 9 shows representative micrographs of single and double coatings on cemented tungsten carbide. The dark regions between the coating and the substrate in these figures are the brittle, η-phase, which should be avoided to improve adhesion and increase the transverse rupture strength (TRS). A very thin (ca 5-μm) coating effectively reduces crater formation on the tool face by one or two orders of magnitude relative to the uncoated tools. At higher speeds, TiC oxidizes and loses its effectiveness. Hence Al_2O_3, which has good wear resistance (both flank and crater wear) at high temperatures, is used. Because TiN is known to provide low friction, it is generally used as the topmost layer. In addition, its lustrous gold color enhances marketability as well as ready recognition of tool wear. Most of the refractory hard coatings (either single or multiple) including carbides, borides, nitrides, oxides, or their combination, were patented. The trend is toward multiple coatings of TiC, TiCN, Al_2O_3, and TiN by a combination of high temperature CVD, medium temperature CVD, and PVD. For example, a multilayer consisting of 10 layers of TiC, TiCN, and TiN and four layers of alumina separated by three layers of TiN has been used (85). This tool showed considerable improvement in flank wear resistance when machining hot-rolled steel, chilled cast iron (54 R_C), and a nickel-base superalloy, Inconel 718, over commercial multilayer alumina-coated tool. Similarly an alumina-coated tool having an initial layer of 3-μm TiC followed by 19 layers of alumina and 19 layers of TiN to a total thickness of 6 μm, performed with improved crater and flank wear when machining an AISI 1060 steel and AISI 1045 steel in both continuous and interrupted cutting (86).

In the CVD process, several thousand tools are loaded in a vacuum chamber and are initially heated to a temperature of ca 950–1050°C for a coating of either TiC or Al_2O_3, or a combination of the two. This initial temperature is generally $<900^\circ C$ for TiN. The high temperatures ensure good interfacial bonding between the

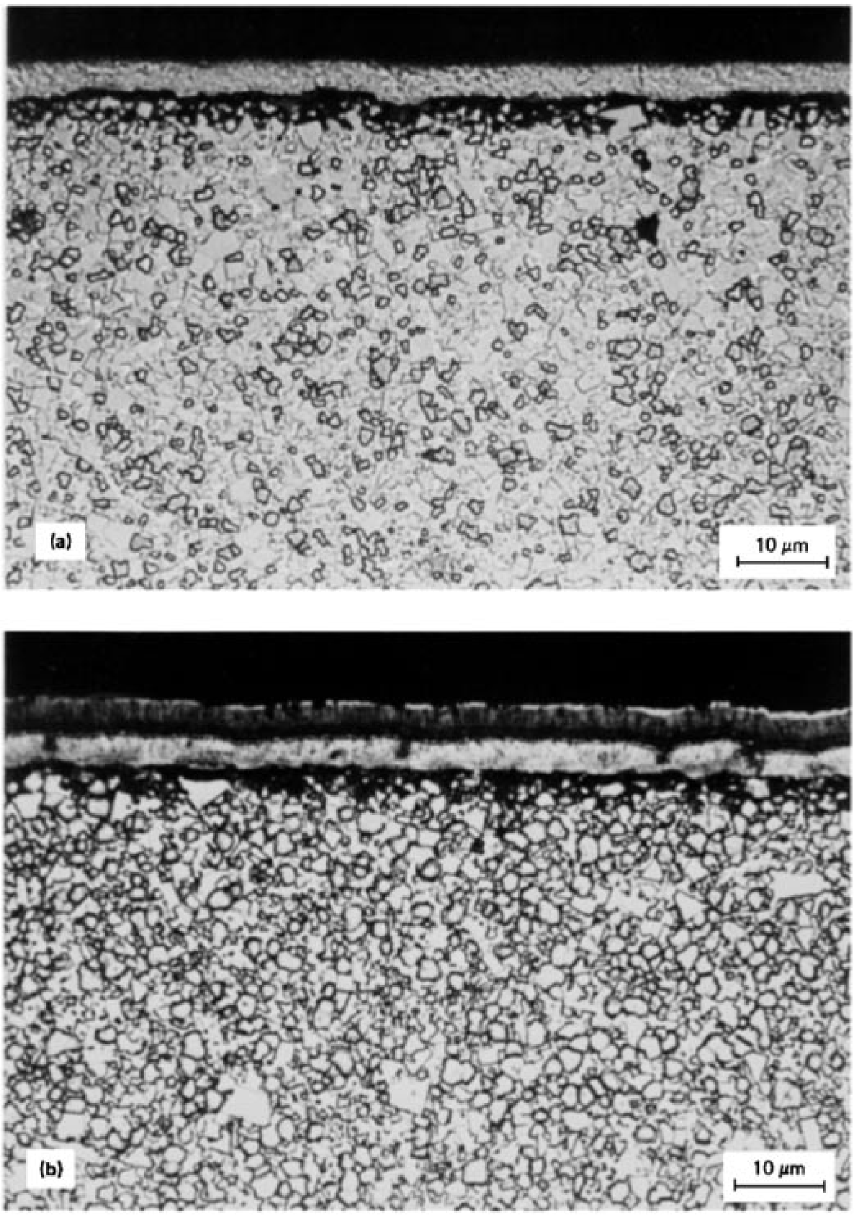

Fig. 9. Representative micrographs of (**a**) single (TiC) and (**b**) double (TiN and TiC) coatings on cemented tungsten carbide (9).

coating and the substrate, provided the material of the substrate is properly tailored for the coating. For a TiC coating, $TiCl_4$ and CH_4 are passed through a reaction chamber containing the tools to be coated in a hydrogen atmosphere at a pressure of ca 101 kPa (1 atm) or less. TiC deposits on the tool face from the vapor phase and establishes a metallurgical bond with the substrate at high temperatures. Alternatively, $TiCl_4$ and carbon (from the cemented carbide substrate) can react in a hydrogen atmosphere. A coating of ca 5–10 μm thickness is deposited for optimum performance, and the process requires ca 8–24 h. In the case of TiC coating, depletion of carbon from the substrate results in the formation of a brittle η-phase, $Co_xW_yC_z$, and associated microporosity at the coating–substrate interface. This brittle η-phase can reduce the transverse rupture strength (TRS) of coated tools by as much as 30% of the substrate's TRS (87), resulting in premature tool failure, especially in interrupted cutting or in roughing. This latter can be a serious limitation, in that interrupted cutting and hence methods to overcome or to augment the strength through other methods become essential for the success of the coated tools. Co-enrichment at and near the rake face of the tool, use of a lower temperature PVD technique which effectively eliminates the formation of the brittle η-phase, and use of a topmost coating of TiN by PVD to introduce residual compressive stresses are some of the attempts to overcome this deficiency. CVD coating technology has reached a stage wherein turnkey systems are available for coating a variety of materials.

Some of the early coated tools were notorious in regard to lack of adhesion of the coating to the substrate, owing to the presence of η-phase at the interface, and consequently exhibited inconsistent performance. This problem has been largely eliminated by a number of process and metallurgical innovations and technological advances resulting in more uniform coatings, better adhesion of the coating to the substrate with minimal interfacial η-phase and associated brittleness. For example, to minimize the affect of η-phase at the interface on coated tool performance, medium (700–900°C) temperature CVD coating technology was developed (70, 71). Using a mixture of $TiCl_4$, H_2, and an organic C–N compound such as acetonitrile, TiCN coatings were developed. Similarly, PVD technology wherein the substrate is not heated above 450°C is applied to cemented carbides to eliminate the interfacial η-phase. Because the coating is extremely thin, the edge of a coated tool should be prepared, eg, by honing a radius or providing a small negative rake land, prior to coating and should not be altered subsequently. This ensures uniform coating around the edge. Reasons for honing include (*1*) honing restores some mechanical strength lost during the CVD process, and (*2*) having minimizes the tendency of extensive η-phase formation. Alternatively, application of coating by the PVD technique overcomes this problem without the need for honing (43). Whenever a sharp cutting edge is required, the latter technique is used, preferably with a submicrometer carbide substrate.

For a coating of Al_2O_3 on cemented multicarbide tools by the CVD process, a gaseous mixture of hydrogen, water vapor, and an aluminum halide such as aluminum trichloride is used in the temperature range of 900–1250°C (79). The water vapor is most conveniently formed by reacting hydrogen with CO_2 in the deposition chamber to form CO and water vapor. H_2 is found to be necessary to ensure oxidation of aluminum at the carbide interface and to form a dense, adherent coating. To enhance adhesion of the coating to the substrate an interlayer of a transition metal which is both a carbide former and an oxide former, such as Ti, Ta, Hf, Zr, or Nb, is selected (88). For example, the tool is first coated with a thin layer of TiC by the CVD process. It is then oxidized to form TiO at the surface, to which Al_2O_3 coating is deposited. Interlayers are thus provided for compatibility between oxides and carbides by a gradual transition or by forming oxi-carbides. In this manner, thermal expansion mismatches and chemical incompatibility are minimized. Similarly, TiN coating is deposited by reacting $TiCl_4$ and N_2 or NH_3 in a hydrogen atmosphere, and TiCN coating is deposited by reacting $TiCl_4$, methane, and N_2 in a hydrogen atmosphere.

To take full advantage of the coating potential, substrates are carefully matched or appropriately altered to optimize properties, resulting in significant gains in productivity. For example, various coatings of Al_2O_3 or TiC, or TiN–TiC or Al_2O_3–TiC, combined with different cemented-carbide substrate materials, provide a range of combination properties of the substrate-coating tailored for different applications. Coated tools with steel-grade substrates (multiple carbides) are still recommended for machining steels, whereas tools with nonsteel grade substrates are recommended for other materials. The coated tools can be used at higher speeds or higher removal rates, or for longer life at the prevailing speeds. However, as multilayer coating process technology advances, the need for large amounts of carbides for steel machining will decrease.

The modes of wear are different on the rake and the clearance faces. Thus coated tool technology has advanced in the direction of selective compositions and modifications of the substrate in strategic areas to prolong the tool life and to make the coated tool more versatile (89, 90). This is especially true in interrupted cutting

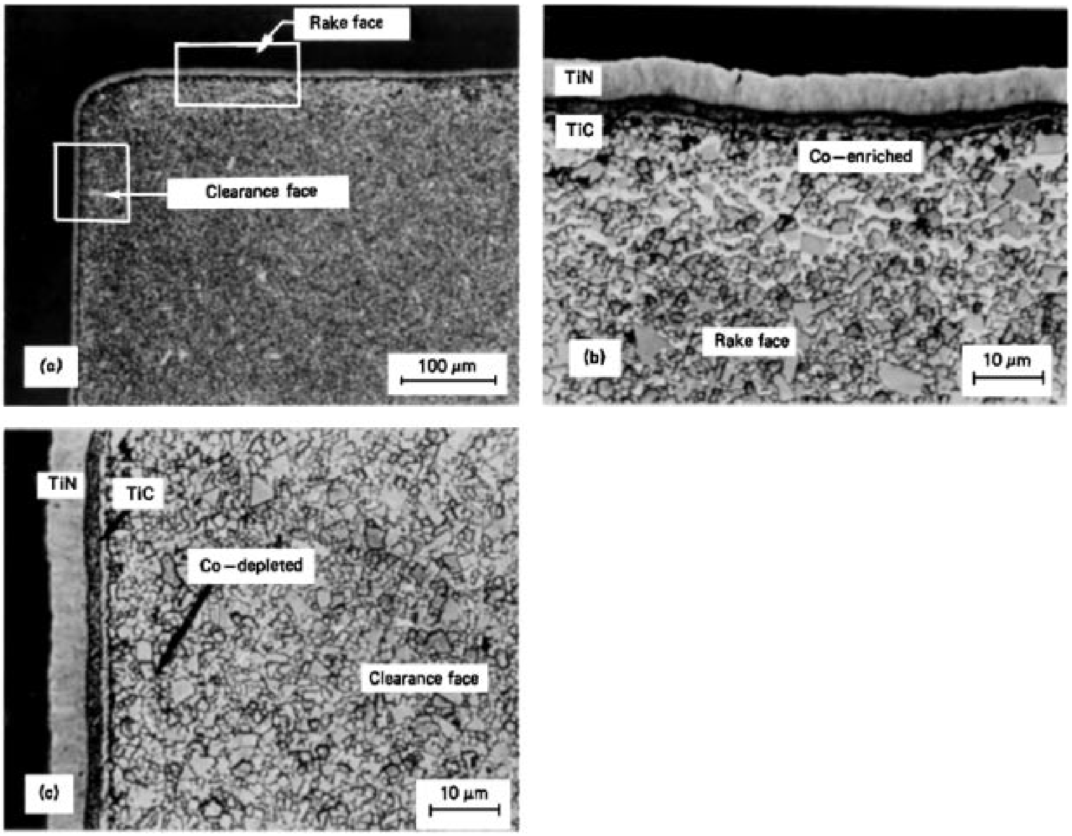

Fig. 10. Optical micrographs of (**a**) a WC tool, and the areas around the (**b**) rake face and (**c**) flank face, showing details of Co enrichment near the rake face and Co depletion near the flank face for the first generation Co enrichment technique.

or heavy-duty cutting, where wear resistance, edge strength, and deformation resistance are all required. An example of the approach in this direction is the development of a multiphase (Ti–C–N or Ti–Al–N)-coated tools with a straight WC–Co-enriched layer (ca 25 μm thick) on the rake face and a Co-depleted layer on the clearance face of a multicarbide (86% WC, 8% (Ti,Ta,Nb)C, and 6% Co) substrate. The Co-enriched zone provides impact resistance characteristic of a high Co-cemented WC, while the flank face depleted of Co provides the superior deformation resistance and high abrasion resistance. This concept was further extended to coatings of TiC–TiCN–TiN on multicarbide with a Co-enriched zone at the rake face, and Co-depleted zone at the flank face provided superior edge while maintaining the edge and crater wear resistance of the coated layers (Fig. 10). The technique used for the Co-enrichment of the tool shown in Figure 10 resulted in a zone near the surface having Co content much higher than desired (200–300% of the bulk binder content instead of 150–200%).

The Co enrichment of the surface layers (25–40 μm) can be achieved either during vacuum sintering or by subsequent heat treatment of the cemented carbides in certain substrates. The Co-enriched zones are characterized by Type A (diameter of the pores ca 10 μm) or Type B (diameter of the pores <40 μm) porosity in the cemented carbide (35, 89, 90). Co-enrichment via heat treatment occurs more readily when the alloy contains a W-lean Co binder with a magnetic saturation in the range of 0.0145–0.0157 $\mathrm{T{\cdot}cm^3/g}$ (145–157 $\mathrm{G{\cdot}cm^3/g}$). Additions of small amounts of TiN along with the necessary carbon to the base powder mix promotes the formation of the required magnetic saturation of the Co binder alloy, which would otherwise be difficult to achieve. This is accomplished by adding ca 0.5–2% of fine-grain TiN or TiCN to the WC–Co alloys. Because TiN is not completely stable during vacuum sintering, partial volatilization of nitrogen takes place, resulting in fine A- or B-type porosity in the material. Typically a layer of Co or C is formed on the surface of the substrate during this process which is removed by grinding prior to coating in order to obtain adherent bonding between the coating and the substrate. This also ensures the appropriate thickness of the Co-enrichment layer (ca 25 μm) on the rake face. As this is a surface-enrichment process all around the insert, the Co enrichment on the flank face of the surface has to be removed by grinding, resulting in the required Co-depleted zone on the flank face, leaving the Co enrichment zone on the rake face. Using this technique, the peripheral Co-enriched zone (ca 10–25 μm wide) is completely devoid of solid solution cubic carbides and the distribution of Co is homogeneous and nonstratified, unlike the first-generation coatings. This development, followed by multilayer coatings of Ti–C–N, or TiC–TiCN–TiN, has enabled users to conduct heavy interrupted machining, such as that encountered in scaled forgings and castings at low cutting speeds (89–91). Further, the introduction of TiC–Al_2O_3–TiN coating on this substrate material has enabled improved impact resistance as well as bulk deformation resistance at high temperatures, thus facilitating high speed interrupted machining (92).

A second technique involves heating the cemented tungsten carbide to the solidus–liquidus temperature region of the binder phase in a decarburizing atmosphere, such as CO_2 gas (93). Decarburization occurs at the surface whereby the carbon concentration at the surface is reduced to reach the solidus line of the binder phase, and the liquid phase solidifies. As a result, the liquid phase is supplied to the inner portion, and this also reaches near the surface where it is decarburized to reach the solidus line and this again solidifies. This procedure is repeated until Co is enriched in the zone near the rake face.

A third method is similar to the first one, in that additions of aluminum nitride in sufficient amounts (5–10%) are made, instead of TiN as in the case of the first method, to the cemented carbide mix and vacuum sintered containing Type B1 porosity. This enhances the surface toughness of the cemented carbide by promoting binder enrichment and depletion of aluminum nitride near the peripheral surface, obtained by decomposition of aluminum nitride during sintering (94). The Co-enriched zone is also characterized by the presence of straight tungsten carbide and depletion of multicarbides. Thus this zone consists of Co-enriched tungsten carbide, and after appropriate coating, the tool would provide the required toughness in interrupted or heavy-duty cutting.

Physical vapor deposition (PVD) technology includes vapor deposition, various types of sputtering, and ion plating (95) (see Thin films). The mean free path in PVD is large, and the vapor species arrive at the substrate without extensive gas collision. Consequently, this is a low temperature, line-of-sight process. Because of lower substrate temperatures, adhesion of the coating to the substrate can be a problem. Also, the rate of deposition, especially using straight sputtering, can be low (on the order of nm/min). However, unlike in CVD, the formation of η-phase is practically eliminated owing to lower substrate temperatures (91). Also, the PVD process, eg, iron plating, can introduce compressive residual stresses beneficial to the adhesion of coating to the substrate and the overall performance of the coated tools (87).

Although coated tools have demonstrated significant performance gains over comparable uncoated tools, up to 3–10 times in certain cases, in the initial stages of their introduction several factors contributed to a less-than-complete acceptance. These include inadequate machine-tool systems (the significant performance gains possible with coated tools are accomplished at higher speeds, at higher removal rates, and with more rigid, high power machine tools; most older machine tools are somewhat limited in this respect); nonuniform and inconsistent performance of some earlier coated tools owing to quality-control problems; limited user knowledge, partly because many small-scale users have less knowledge on the performance and application of these coated tools, and partly because the technology is advancing rather rapidly; slightly higher cost, which should not be a consideration because the cost of coated tools is only fractionally higher than that of uncoated tools; larger inventory of different tool grades; and slightly lower toughness reported with some coated tools.

Coated tools, originally developed exclusively for machining steels, are used for a wide range of materials, including various types of steels, cast iron, stainless steel, nickel and Co-base superalloys, and titanium alloys.

The tool manufacturers often encounter a challenge between developing a general-purpose grade that covers a wide range of work materials, cutting conditions, and machining operations to reduce the number of grades the user can stock vs the coatings targeted for niche areas, eg, coatings for machining titanium alloys and coatings for interrupted cutting. Generally, a coating for a niche area is developed first because of the specific need in that area. Subsequently its scope is broadened to cover either a class of work materials, a class of manufacturing operations, or a range of machining conditions. The selection of an appropriate tool material and proper tool geometry, chip groove geometry, coating, and substrate properties for machining a given material is not a simple selection based on rule of thumb, but rather is a more sophisticated decision-making process based on detailed knowledge and experimentation. Of the carbide tools used as of the mid-1990s, some 65% are coated, especially those in the United States, Western Europe, and Japan. This percentage is expected to increase considerably (as high as ca 80%) into the twenty-first century.

A thin (ca 5-μm) coating effectively reduces crater formation on the tool face by a factor of two to three relative to uncoated tools. Multiple coatings enable greater (ca 10–15 μm) thickness, create a strong metallurgical bond between the coating and the substrate, provide protection for different work materials, and prolong tool life. Single or multiple coatings of different thicknesses of TiN, TiN–TiCN–TiN, TiN–TiCN, TiN–Al_2O_3–TiC–TiCN, TiN–TiCN–TiC, TiN–Al_2O_3–TiCN, Al_2O_3–TiC, Al_2O_3–TiC–TiCN, TiN–Al_2O_3–TiC, and TiN–Al_2O_3–TiN–Al_2O_3–TiN–Al_2O_3–TiCN engineered on selected cemented carbide (different Co content and straight WC or multicarbide) as well as Co-enriched substrates exist. In addition, one grade of TiN–Al_2O_3 on a Si_3N_4 substrate is also available. These coatings can be combined with suitable chip-groove geometry to result in an optimum level of performance. Coatings on cutting tools have become customized for various work-materials and cutting applications. The trend is toward coating the first layer of TiN or TiCN on the substrate by PVD to minimize or eliminate the formation of η-phase ($Co_xW_yC_z$), as well as to coat the last layer by PVD to induce residual compressive stresses on carbide substrates. Although PVD and medium temperature CVD coatings are beginning to appear, the majority of the coatings on cemented carbides are made by the CVD technology. Technology to coat crystalline Al_2O_3 by PVD is not available as of this writing (ca 1997).

Other coatings, such as TiAlN (96), TiCN, ZrO_2, and ZrN (97), and CrN (98) were developed for special applications. The last was developed for higher speed machining of titanium alloys. Sometimes a coating is developed not for its wear-resistance but for its heat insulation. The case in point is alumina coating of cBN to reduce the heat conductivity at the surface so that the cBN performance can be enhanced (99).

4. Multiple Nanolayered Coatings

The refractory hard materials used for coatings on cutting tools are generally brittle and hence not tough. The fracture mechanism consists of crack initiation at stress concentrations and its rapid propagation to failure (see Fracture mechanics). By arresting the propagation of the cracks, it is possible to increase the toughness of these hard coatings significantly without compromising on hardness. This is accomplished by applying multiple nanolayer coatings of alternating hard and tough materials (see Nanotechnology (see Supplement)). Investigations have been directed at improving the properties of materials significantly by reducing the microstructural or spatial scale of a material system to nanometer dimensions (100, 101). In this approach, a crack initiated in any hard layer is stopped when it reaches the tough layer, facilitating higher toughness (100–103). The number of nanolayers can be several hundred in contrast to the few layers used on cutting tools prepared by CVD techniques.

Nanolayer coatings are generally expected to be harder, tougher, and chemically more stable than coatings of several micrometers or of bulk materials. For example, B_4C or SiC ceramics have high hardness but are not used as cutting tools either in monolithic form or as micrometer-thick coatings because of the ease of oxidation, reaction with most ferrous materials, and more importantly their inherent brittleness. However, in nanolayer coatings in multiples (literally hundreds) of layers having alternating hard material and tough ductile metal, the material is no longer brittle and even if a layer oxidizes after performing its cutting action, the next layer is ready to take its place during subsequent contact with the work-material. This case is somewhat similar to the self-sharpening action of abrasives in grinding, where the abrasives (qv) release new cutting edges by micro cleavage once the previous edges are worn out.

In addition to coatings of alternating hard and tough materials, multiple nanocoatings can also be developed based on alternating hard and lubricating materials, such as a carbide/oxide/boride/nitride–MoS_2 sequence to improve tool life and at the same time decrease friction at the chip–tool interface. Multiple coatings of alternate WC (1.2-nm) and Co (0.8-nm) (104), nanolayered coatings of Al–Al_2O_3 (105, 106), and Ti–TiN coatings (107) have been developed, as have multinanolayer composite coatings consisting of alternate solid lubricant and a metal (108, 109). Solid lubricant–metal multilayer nanocoatings have been developed for tribological applications (110, 111). TiN-based coatings have been introduced to investigate the *in situ* solid lubrication of TiO_2 layers, believed to form by the reaction of TiN with water vapor in the air (112), Ti–TiN, Al–AlN nanolayers for tribological and corrosion protection have been investigated (113). The role of tribology in metal cutting has been discussed (114).

The number of material systems that can be used for nanolayer coatings is virtually unlimited. Any refractory hard material can be used as the hard material; compatible metal can be used as the tough material. Examples of material systems for nanocoatings include the following:

In situ formation of oxidation-protective and low friction layers	
hard carbide/hard carbide systems	B_4C–SiC, HfC–B_4C, HfC–SiC
hard carbide/metal co-sputtered or layered systems	B_4C (HfC, SiC)–W (Al, Cr, Ti, Si, Mg, Zr)
Low friction, low stress coatings	
layered lattice (solid lubricant)/metal systems	MoS_2–Mo, MoS_2–Ag–Mo, TaS_2–Ta, WS_2–W
Low friction, hard coatings	
hard carbide, oxide, nitride, boride/layered lattice (solid lubricant) systems	

Multiple nanolayer coatings are deposited by PVD; chiefly magnetron sputtering using multiple targets is employed. The tool is rotated with respect to the targets. The tool then sees different sputtering targets alternately. Consequently, the coatings are endowed with the benefits associated with the PVD technique, ie, lower substrate temperatures, virtual elimination of the η-phase in the case of carbide substrates, and residual compressive stresses. The sputtering rate and the rotational speed of the tool (or duration during which the tool is exposed to a given target material) determines the thickness of each coating layer. Whereas each layer in the coating is only a few nanometers, the total thickness can be in the range of 2–5 μm. As of this writing (ca 1997), no commercial nanocoatings for cutting tool applications are available in the marketplace. This technology is expected to become widely used in the twenty-first century.

5. Ceramics

Ceramics (qv), one of the newest classes of advanced tool materials, are used on the one hand for high speed finishing operations involving light feeds and on the other for high removal-rate machining involving low speeds and large depths of cut of some difficult-to-machine steels and cast irons (115–119). The ceramics used initially were predominantly alumina based, although silicon nitride-based materials (also called nitrogen ceramics) have been found to be very attractive for high speed machining of gray cast iron (1500 m/min) and nickel-base superalloys (200 m/min or higher) (see Advanced ceramics). Ceramics, in general, are harder, more wear-resistant, more highly refractory, and chemically more stable than cemented carbides and HSS. Notching at the depth of cut line (DCL notching), microchipping of the tool edge, chipping owing to thermal or mechanical cyclic stresses during interrupted cutting, and gross fracture of the tool are the predominant modes of wear experienced with ceramic tools. Because of poor thermal and mechanical shock resistance, interrupted cutting is especially severe on ceramic tools, owing to repeated entry and exit of the cut. Both the tool and the part to be machined must be fully supported, and to prolong life, the machine tool must be extremely rigid. Ceramics are machined either dry or with a heavy stream of coolant, because intermittent application of coolant can cause thermal shock leading to fracture. High speed, high power precision machine tools are desirable to take full advantage of the potential of ceramic tools.

Although ceramic tools were considered for certain machining applications as early as 1905, transverse rupture strength (TRS) under the conditions of cutting was inadequate and the performance inconsistent (115, 116). In the mid-1950s, ceramic tools were reintroduced for high speed machining of steels and gray cast iron for the automobile industry, and slow speed, high removal-rate machining of extremely hard (and difficult to machine) chilled cast iron or forged steel rolls used in the steel industry. These were basically fine-grain ($<5\ \mu$m), alumina-based materials, alloyed with suboxides of Ti or Cr to form solid solutions, and contained small amounts of magnesia as a sintering aid (115). Carboloy in the United States developed an alumina–TiO ceramic (Grade O-30) that is characterized by a grain size of ca 3 μm and reasonably uniform microstructure. The TiO constitutes ca 10%. A density of ca 90–95% theoretical, a hardness of ca 93–94 HRA and a TRS >550 MPa (ca 80,000 psi) were achieved with cold-pressing followed by sintering (120, 121). Similarly, the Carborundum Company developed a nearly pure alumina (Stupalox or CCT 707) having minor additions of MgO as a sintering aid and a grain-growth inhibitor (115). The Vascoloy Ramet/Wesson Corporation manufactured a similar material (VR 97), originally developed by the Norton Company. Extremely rigid, high powered (up to 450 kW (600 hp)) machine tools were specially designed having high stiffness and high precision, enabling material removal rates of two to three orders of magnitude higher than for conventional machine tools (122). Similarly, high (up to 5000 rpm) speed machine tools were specially built for machining gray cast iron using full automation to take advantage of the potential of this tool material.

Several factors have rejuvenated interest in the development and application of ceramic cutting tools (115–119). Applications of advanced ceramics for structural applications, advances in the ceramic-processing technology; progress in the understanding of the toughening mechanisms in ceramics; rapidly rising manufacturing costs; the need to use materials that are increasingly more difficult to machine; rapidly increasing costs and decreasing availability of W, Ta, and Co, which are the principal and strategic raw materials in the manufacture of cemented-carbide tools; and advances in machining science and technology have all played a role.

A comparison of the physical properties of ceramic tools and carbide tools is given in Table 11. Ceramics are harder (hence, more abrasion-resistant), have a higher melting temperature (thus are more refractory), and are chemically more stable up to their melting temperatures. Ceramics are, however, less dense and less tough (lower TRS), have lower thermal conductivity, and have lower thermal expansion coefficients than cemented carbides. Toughness of ceramics having smaller grain size can be improved by the introduction of a more ductile second phase. Because of lower TRS and high refractory characteristics, ceramics are generally recommended for higher cutting speed ($\geq 300-\mathrm{m/min}$), a lower rate of material removal, ie, high speed finish machining, and continuous-cutting applications. Lower fracture toughness (the value for alumina is ca 2.3 $\mathrm{MPa{\cdot}m^{1/2}}$) is the main limitation in the application of ceramics for heavy or interrupted cutting.

The next advancement in alumina-based ceramics is the development of pure alumina and alumina–TiC dispersion-strengthened ceramics (117). Alumina–TiC-based ceramics contain ca 30 wt % TiC and small amounts of yttria as a sintering agent, resulting in a density close to 99.50% theoretical. High purity, fine grain size, and elimination of porosity are the principal reasons for the high TRS (700–900 MPa (ca 100–130 ksi)). The

Table 11. Physical Properties of Ceramic and Cemented-Carbide Cutting Tools[a]

Property[b]	Ceramics	Cemented carbide[c]
hardness, HRA[d]	91–95	90–93
TRS[e] for alumina-based ceramics, MPa[f]	690–930	1590–2760
melting range, °C	ca 2000	ca 1350
density, g/cm^3	3.9–4.5	12.0–15.3
modulus of elasticity, E, GPa[g]	410	70–648
grain size, μm	1–3	0.1–6
compressive strength, MPa[f]	2760	3720–5860
tensile strength, MPa[f]	240	1100–1860
thermal conductivity, W/(m·K)		41.8–125.5
thermal expansion coefficient, $10^{-6}/°C$	7.8	4–6.5

[a] Ref. 9.
[b] The exact properties depend upon the materials used, grain size, binder content, volume fraction of each constituent, and processing method.
[c] Coated carbides are not included.
[d] Rockwell hardness A scale.
[e] Transverse rupture strength.
[f] To convert MPa to psi, multiply by 145.
[g] To convert GPa to psi, multiply by 145,000.

slightly more expensive hot-pressed materials usually have higher TRS and more consistent performance than cold-pressed materials. The fracture toughness of this material is ca 3.3 $(\mathrm{MPa \cdot m})^{1/2}$). The fracture toughness and TRS are improved through obstruction of the cracks, crack deflection, or crack branching caused by the dispersed hard TiC particles. The lower thermal expansion coefficient and higher thermal conductivity of the composite compared to straight alumina increases the thermal shock resistance and thermal shock cycling experienced in interrupted cutting. At temperatures exceeding 800°C, TiC, and TiN begin to oxidize and lose their strengthening properties. Figure 11**a** shows a micrograph of a hot-pressed alumina–TiC ceramic. Other Al_2O_3-based ceramic tools include Al_2O_3–TiB_2, Al_2O_3–ZrO_2, and silicon–aluminum–oxygen–nitrogen (Si–Al–O–N). The latter two materials are slightly less hard than alumina or Al_2O_3–TiC, but significantly tougher. When machining superalloys, hard chill-cast irons, and high strength steels in the medium speed range (ca 150-m/min), longer tool life results because of lower flank wear and, more important, lower DCL notching.

An Al_2O_3–ZrO_2 ceramic (Cer Max 460) was introduced in the United States by Carboloy. A similar material performs exceptionally well in the grinding industry as a tough abrasive in heavy-stock grinding operations, such as cut-off and snagging. The high toughness and superior grinding performance of this alloy are attributed to the rapid freezing of the alloy from the melt, which results in a dendritic freezing of the eutectic structure. Alternatively, a fine dispersion of unstabilized ZrO_2 in the matrix of Al_2O_3 can give rise to stress-induced transformation of tetragonal zirconia particles, inducing microcracks which absorb elastic energy and prevent cracks from propagating, thereby increasing the fracture toughness. The three popular compositions contain 10%, 25%, and 40% ZrO_2, respectively; the remainder in each case is alumina. The 40% ZrO_2–Al_2O_3 composition is close to the eutectic. The higher ZrO_2 compositions are less hard, but tougher.

The micrograph of a fracture surface of an alumina–zirconia alloy is shown in Figure 11**b**. The zirconia particles are concentrated predominantly at the alumina–grain boundaries. Although the fracture is intergranular, the presence of these particles is believed to provide additional toughness before failure can occur by fracture. In some machining tests on a tough chill-cast iron (HRC 42–44) used for steel rolls, this material performed exceptionally well, showing very little wear in plunge cuts at 150 m/min cutting speed, a feed rate of 0.4 mm per revolution, and a width of cut of 25.4 mm over a straight-alumina tool. Chipping was the predominant mode of wear in the latter case. Similarly, when machining solution-treated and aged Inconel 718, a nickel–iron base superalloy (HRC 42–44), at a cutting speed of 150 m/min, a feed rate of 0.3 mm per revolution, and a

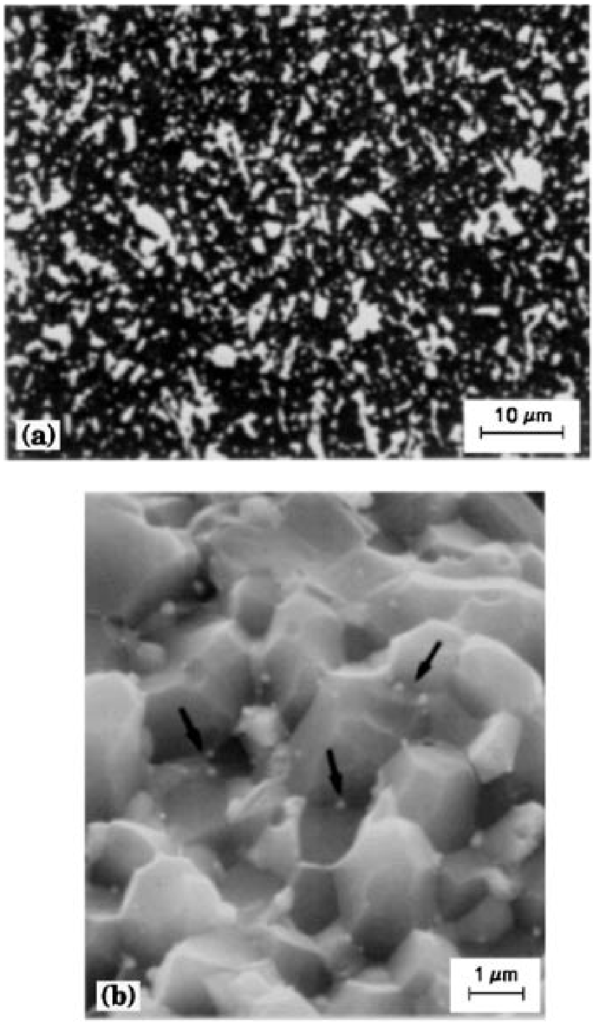

Fig. 11. Micrographs of (**a**) a hot-pressed alumina–TiC ceramic showing a white TiC phase and a dark alumina phase (3) and (**b**) a fracture surface of an alumina–zirconia alloy precipitation of zirconia at the alumina grain boundaries (3).

depth of cut of 0.3175 mm, this material (2.54-cm round) gave a tool life of >8 min and yielded an excellent finish (1–2.5 mm) on the machined surface. These tests were conducted dry. Based on other successful high speed machining tests (ca 300 m/min) where a coolant-lubricant was used, higher tool life at the same speed or increased cutting speed for the same tool life was obtained.

The second interesting class of ceramic tool material under development is based on Si_3N_4, either nearly pure Si_3N_4 (except for some minor additions of sintering aids) or having various additions of aluminum oxide, yttrium oxide, and TiC (123–134). It is a spin-off of the high temperature–structural ceramics technology developed in the 1970s for automotive gas turbines and other high temperature applications. Ford Motor Co. developed a ceramic tool of Si_3N_4 having additions of ca 12% yttria (Grade S 8). Norton Co. developed a Si_3N_4 ceramic based on MgO but has never commercialized it for cutting tool applications. Instead, it concentrated on advanced structural applications, including hybrid ceramic bearings. Similarly, General Electric and

Westinghouse also developed Si_3N_4-based ceramics for high temperature–structural applications but did not extend them to cutting tools.

Si_3N_4 is a covalently bonded material that exists in two phases, α and β. The structure of these phases is derived from the basic Si_3N_4 tetrahedra joined in a three-dimensional network by sharing the corners, with each nitrogen corner being common to the three tetrahedra. This material provides a number of favorable properties, including higher elevated temperature strength, thermal stability, low thermal expansion coefficient, higher thermal conductivity, and higher fracture toughness than alumina. But because of the low self-diffusion coefficient, it is virtually impossible to fabricate pure Si_3N_4 into a dense body with no porosity by conventional sintering or hot pressing techniques without the sintering aids. The predominant impurity in Si_3N_4 is SiO_2, which is present on the surfaces of the Si_3N_4 particles. In the synthesis of Si_3N_4, sintering aids, such as MgO, Y_2O_3, TiO, and Cr_2O_3, are combined with fine powder of α-Si_3N_4, ball-milled, and used as the starting material for consolidation. It is cold-pressed to shape and hot-pressed in a N_2 atmosphere at temperatures ca 1600°C. At the densification temperature, the sintering aids combine with SiO_2 to form a glassy liquid phase. α-Si_3N_4 particles dissolve in this liquid phase and precipitate out as β-Si_3N_4. The β-Si_3N_4 nuclei grow as elongated grains and form an interlocked grain structure. The higher fracture toughness (4–6 $MPa{\cdot}m^{1/2}$) and higher strength of Si_3N_4 are attributed to this elongated grain structure. Commercial Si_3N_4 is characterized by a two-phase structure, consisting of β-Si_3N_4 crystallites and an intergranular bonding phase.

Si_3N_4 is marketed by most tool manufacturers and used extensively in high speed machining of gray cast iron. However, to take full advantage of this tool material at high speeds, more rigid, high speed, high power machine tools are required. Many automotive and other industries are working with the machine-tool builders of these special machine tools. The fracture toughness of pure Si_3N_4, higher than alumina, is not adequate for rough machining, interrupted cutting, or in machining of castings with irregular surfaces or with a scale. To address this problem, GTE developed a ceramic tool of Si_3N_4 having additions of yttria (ca 6%), alumina (ca 2%), and a fine dispersed phase of TiC (ca 30%) (Grade Quantum 5000). Other additions of the dispersed phase include TiN, HfC, or a combination of TiN and TiC, which increases the hardness and fracture toughness (via crack interactions with the dispersoid and crack deflection) of the composite material. Many tool manufacturers have similar grades. Similarly, SiC or Si_3N_4 whisker-reinforced Si_3N_4 was developed to increase the fracture toughness of the base material for interrupted cutting applications.

In the late 1970s Lucas Industries Ltd. of the United Kingdom developed a ceramic tool of Si–Al–O–N with additions of yttria, and marketed under the trademark SYALON (in the United States this material was marketed initially as KYON by Kennametal Inc. under license; Al_2O_3-based materials are sold as KYON as of 1997). Oxygen (O^{2-}) can be substituted for nitrogen (N^{3-};) in the β-Si_3N_4 crystal provided Al^{3+} is simultaneously substituted for Si^{4+} to maintain charge neutrality. SiAlON tools are produced by sintering. The powder charge consists of a mixture of Si_3N_4, AlN, Al_2O_3, and Y_2O_3. The last is added as a sintering aid for full densification. The powder mix is first ball-milled, dried, pressed to shape by cold-pressing, and subsequently sintered at a temperature of ca 1800°C under isothermal conditions for ca 1 h before it is allowed to cool slowly. Yttria reacts with β-Si_3N_4 to form a silicate which is a liquid at the sintering temperature. The resulting Si_3N_4 material thus has a glassy intergranular phase. Some properties of the Sialon material are given in Table 12 (135).

Further developments in microstructure and composition may yield an even more refractory material consisting of β-Si_3N_4 and an intergranular phase of yttrium aluminum garnet (YAG) without an intergranular glassy phase. Similar to the alumina–zirconia ceramic, this material offers significant improvements in tool life, consistency in tool performance (more reliability), and higher removal rates possible at reasonable cutting speeds (90–125 m/min) when machining nickel-base superalloys. With the increasing trend toward computer-controlled machining, consistency and reliability of tool performance are crucial. Furthermore, the trend toward more than one machine tool per operator is resulting in lower and more manageable but reliable cutting speeds. SiAlON, alumina–TiC, alumina–zirconia, and straight alumina are some of the tool materials that might meet the needs of these trends.

Table 12. Composition and Properties of Sialon Material[a]

Parameter	Value
Composition	
Si_3N_4, wt %	77
Al_2O_3, wt %	13
Y_2O_3, wt %	10
Properties	
density, g/cm^3	3.2–3.4
hardness, GPa[b] (kgf/mm^2)	17.65
Young's modulus, GPa[b,c]	300
compressive strength, MPa[d]	>3500
thermal conductivity, W/(m·K)	20–25
thermal expansion coefficient, $10^{-6}/°C$	32

[a] Ref. 135.
[b] To convert GPa to psi, multiply by 145,000.
[c] Value is equal to 1800 kgf/mm^2.
[d] To convert MPa to psi, multiply by 145.

Because of its high toughness and good thermal shock resistance, test results indicate the possibility of using square-, triangular-, and diamond-shaped tools of SiAlON for machining superalloys in the intermediate speed range (ca 150 m/min) where only round tools are used currently with other ceramics.

Even though Si_3N_4 and SiAlON tool materials are used extensively for high speed machining of cast iron and machining of nickel-base superalloys, respectively, they could not be used for machining steels. To take advantage of their improved fracture toughness as well as their ability to machine ductile C-1, ceramic coatings on Si_3N_4, SiAlON, and modified compositions of the two were developed. Thin (2–5-μm) coatings on monolithic ceramic substrates have been developed mainly to limit chemical interactions between the tool and the steel work materials. Also, to take advantage of the high temperature deformation resistance of this material and to minimize chemical interactions when machining steels at high speeds, single (TiC, TiN, AlN, Al_2O_3) or multiple coatings of TiC–TiN, or Al_2O_3–TiC on silicon nitride, SiAlON, and SiAlON dispersed with TiC substrates were developed that are similar to the coatings on cemented carbides (136–140) (Fig. 12). Whereas attempts are continuously being made to improve the adhesion between the coating and the substrate, the problem remains. Even if this problem were to be solved, the extent to which such coated ceramic tools are used vs competing materials for high speed machining of steels and other materials would depend on the need for them and the economics of machining. The coated ceramics are still in the experimental stage and (ca 1997) are not available commercially.

A newer whisker-reinforced ceramic composite (SiC whisker-reinforced alumina) material possessing improved fracture toughness (K_{1C} ca 8.5–9 $(MPa{\cdot}m)^{1/2}$) has been introduced as a cutting tool for machining of nickel–iron-base superalloys used in aircraft engines (141–154). Single crystal whiskers of SiC possess very high tensile strength (7 GPa (1×10^6 psi)). SiC also has higher thermal conductivity and higher coefficient of thermal expansion than alumina. Consequently, the composite exhibits higher strength, fracture toughness, and thermal shock resistance.

SiC whiskers are commonly made from rice hulls, a waste product of agriculture (150). Rice hulls have a high (15–20%) ash content relatively high in silica (>95%) and cellulose. Thermal decomposition of this very high surface area material provides intimate contact in the rice hulls and SiC whiskers are readily formed. In the preparation of the SiC whisker-reinforced composite material, SiC whiskers (0.5–1 μm in diameter and 10–80 μm long) about 20% volume fraction are mixed homogeneously with micrometer-sized fine powder of alumina. The mixture is then hot-pressed to over 99% of the theoretical density at a pressure in the range of 28–70 MPa and temperature in the range of 1600–1950°C for pressing times varying from about 0.75 to 2.5 h.

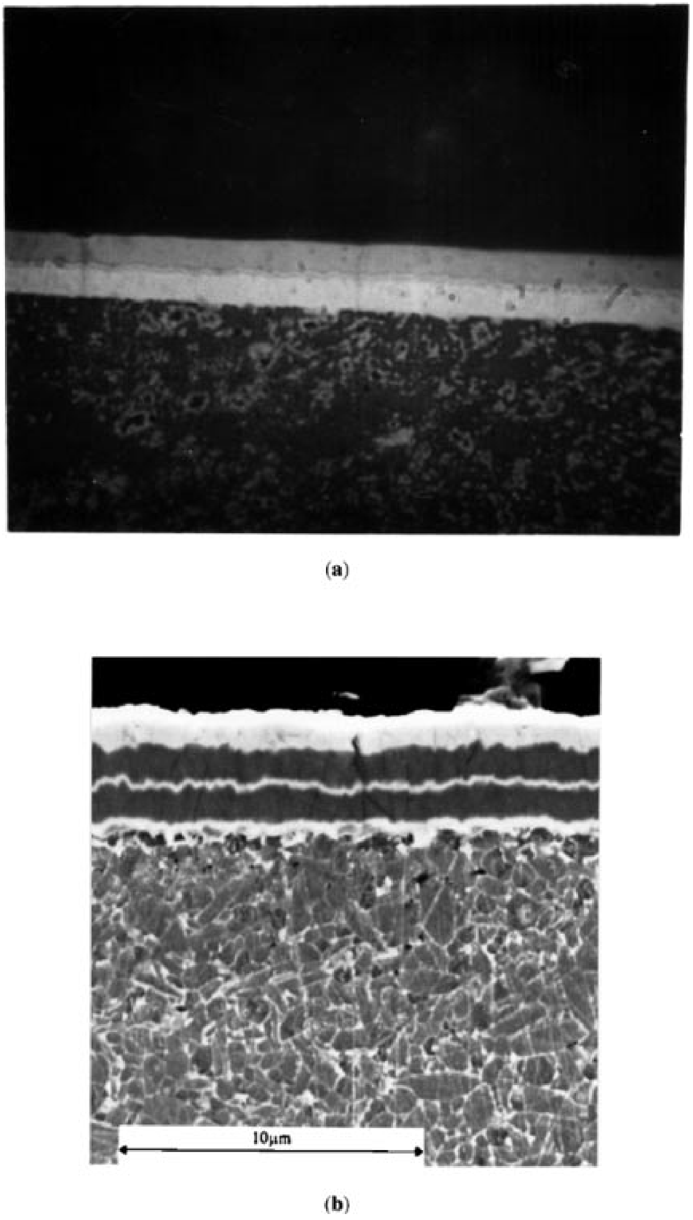

Fig. 12. (**a**) A cross section of multiple coatings of TiN on TiC on a silicon nitride-based tool material; (**b**) multicoatings on a SiAlON-based tool material.

The fracture toughness of this SiC whisker-reinforced alumina material is by far the highest among ceramic cutting tools, nearly twice that of its closest ceramics, Si_3N_4 and Sialon. Details of the micromechanisms for improved fracture toughness of this ceramic composite tool material have not been clearly established (151, 152).

Although monolithic Si_3N_4 is a reasonably tough ceramic tool material (K_{1C} ca $4.7\ MPa{\cdot}m^{1/2}$) compared to alumina, its fracture toughness is not high enough for interrupted cutting (milling or rough turning) of cast iron. Thus a tougher silicon nitride tool reinforced with SiC or Si_3N_4 whiskers was developed for this application. For example, a 30 vol % SiC whisker-reinforced Si_3N_4 showed an increase of 40% in the fracture toughness (K_{1C} of ca $4.7\ MPa{\cdot}m^{1/2}$ for Si_3N_4, compared to ca 6.4 $(MPa{\cdot}m)^{1/2}$ for SiC whisker-reinforced material) and 25% increase in strength (TRS) (151). Similarly, to increase the toughness of the matrix material even further, SiC whisker-reinforced Al_2O_3–ZrO_2 material was developed (145).

The SiC whisker-reinforced alumina composite, a model for engineered materials, has opened new vistas for tool material development. Whereas SiC whisker-reinforced alumina is used extensively for the machining of nickel-base superalloys, SiC whiskers react chemically with steel, causing rapid wear on the rake face. Attempts are underway to replace SiC whiskers with less reactive whiskers such as TiC or TiN.

Ceramic tools are inherently more brittle than cemented carbides, and a tool geometry of -10 deg rake and $+10$ deg clearance is recommended instead of -5 deg rake and $+5$ deg clearance (for cemented carbides). In interrupted cutting, attempts should be made to shift the point of application of the load away from the cutting edge to minimize chipping. Suitable edge preparation involving honing a small radius or a small negative land on the rake face is also recommended. The work materials recommended include hardened steels, chill-cast iron, and superalloys (Ni-base and Co-base).

Certain ceramic tools, especially those based on alumina, are not suitable for machining aluminum, titanium, and similar materials because of a strong tendency to react chemically. They are also not generally suited for low speed and intermittent cutting operations because of failure by chipping, unless they are used on extremely rigid high precision machine tools. Poor thermal shock resistance prevents the intermittent application of cutting fluids. Hence either heavy flooding or no coolant at all is recommended for machining with ceramic tools.

6. Diamond

Diamond is the hardest (Knoop hardness ca 78.5 GPa (ca 8000 kgf/mm^2)) of all known materials. Both the natural (single-crystal) and synthetic (polycrystalline sintered body) forms can be used for cutting-tool applications. Diamond tools exhibit high hardness, good thermal conductivity, ability to form a sharp edge by cleavage (especially the single-crystal natural diamond), low friction, nonadherence to most work materials, ability to maintain a sharp edge for a long period of time, especially when machining soft materials like copper and aluminum; and high wear resistance. Sometimes if the surface of a tool material is somewhat rough, metal may be stuck in the valleys of the tool surface and subsequent buildup can occur between this metal and the chips. Disadvantages of diamond tools include extensive chemical interaction with metallic elements of Groups (4–10) (IVB–VIII) of the Periodic Table (diamond wears rapidly when machining or grinding mild steel; it wears less rapidly with high carbon alloy steels than with low carbon steel and is occasionally employed to machine gray cast iron (high carbon content) with long life); a tendency to revert at higher (ca 700°C) temperatures to graphite and oxidize in air; extreme brittleness (single-crystal diamond cleaves easily); difficulty in shaping and reshaping after use; and high cost.

Steels account for a significant fraction of the work materials machined. Thus the inability to use diamond for machining steels is a significant limitation where general machining is concerned. There are, however, other applications where diamond is the ideal material, such as for ultraprecision machining of aluminum and copper for laser mirror applications where it is almost impossible to use other tool materials to produce such a surface to provide a long tool life economically. Use of a single-crystal diamond as microtome knives is another unique application. High quality, single-crystal industrial diamonds are the tools of choice for these applications because of long life and ability to machine accurately (with an extremely sharp edge formed by cleavage). Lower quality industrial diamonds are extensively used in high speed machining of

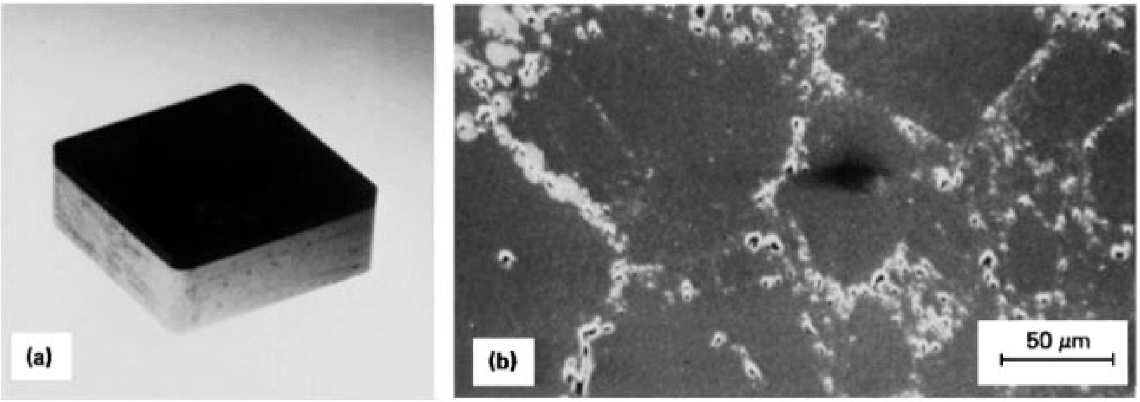

Fig. 13. (**a**) Photograph of a polycrystalline diamond tool showing a thin layer (ca 0.5–1.5 mm) of fine-grain size diamond particles sintered together and metallurgically bonded to a cemented carbide base; (**b**) micrograph of the polycrystalline diamond tool showing extensive diamond-to-diamond bonding.

aluminum–silicon alloys in the automobile industry; in polymers and glass–epoxy composites in the aircraft industry; in copper commutators in the electrical industry; for machining nonferrous (brass, bronze) and nonmetallic materials; for cold-pressed sintered–carbide performs for the metal-cutting and metal-forming industries; to shape and cut stone and concrete; and as dressing tools for alumina grinding wheels. Industrial quality natural diamonds give unreliable performance, caused by easy cleavage and unknown amounts of impurities and imperfections. Regrinding of these tools is difficult and expensive.

Limited supply, increasing demand, and high cost have led to an intense search for an alternative, dependable source of diamond. This search led to the high pressure (ca 5 GPa (0.5×10^6 psi)), high temperature (ca 1500°C) (HP–HT) synthesis of diamond from graphite in the mid-1950s (153–155) in the presence of a catalyst–solvent material, eg, Ni or Fe, and the subsequent development of polycrystalline sintered diamond tools in the late 1960s (156).

The polycrystalline diamond tools consist of a thin layer (ca 0.5–1.5 mm) of fine grain-size particles sintered together and metallurgically bonded to a cemented-carbide base. The cemented carbide provides the necessary elastic support base for the hard and brittle diamond layer above it (Fig. 13**a**). These tools are formed by a HP–HT process at conditions close to those used for the synthesis of diamond from graphite. Fine diamond powder (1–30-μm) is first packed on a support base of cemented carbide in the press. At the appropriate sintering conditions of pressure and temperature (in the diamond-stable region), complete consolidation and extensive diamond-to-diamond bonding takes place (Fig. 13**b**). Stress concentration at the sharp corners of the diamond crystals during sintering subjects these areas to local stresses of perhaps an order of magnitude higher than nominal (ca 50 GPa (5×10^6 psi)). As a result, individual diamond crystals are work-hardened (154), resulting in a sintered diamond compact which is probably much harder than an undeformed diamond, and consequently the abrasion resistance of the tool is increased. In addition to diamond-to-diamond bonding, good metallurgical bonding is established between the diamond layer and the cemented-carbide support base in this process. However, some of the binder phase and other impurities from the underlying cemented carbide can diffuse into the diamond layer above it. This can affect the high temperature performance of the polycrystalline diamond. In order to eliminate this effect, the metallic impurities in the polycrystalline diamond are leached and the resulting diamond layer is subjected to HP–HT conditions to obtain a much denser, impurity-free material with superior performance (157). The polycrystalline diamond tools are then finished to shape by laser cutting or electrodischarge machining (EDM), followed by grinding, polishing, and lapping to size, finish, and accuracy.

Sintered polycrystalline diamond tools of various grain sizes are fabricated in an assortment of shapes (squares, rounds, triangles, and sectors of a circle of different included angles) and sizes from round blanks. The main advantages of sintered polycrystalline tools over natural single-crystal tools are better control over inclusions and imperfections, higher quality, and greater toughness and wear resistance (resulting from the random orientation of the diamond grains and the corresponding lack of simple cleavage planes). In addition, the availability of sintered diamond tools is not dictated by nature or some artificial control; thus, such tools can be manufactured to meet strategic needs.

Sintered polycrystalline diamond tools are much more expensive than conventional cemented-carbide or ceramic tools because of the high cost of the processing technique and the finishing methods used. Diamond tools, however, are economical on an overall-cost-per-part basis for certain applications because of long life and increased productivity.

Sintered diamond tools are used for applications similar to the lower quality industrial diamonds. Because of high reactivity, they are not recommended for machining soft low carbon steels, titanium, nickel, cobalt, or zirconium. Because they are inherently brittle, they are used with a negative rake (-5°) geometry with suitable edge preparation on materials that are difficult to machine, such as pressed and sintered cemented tungsten carbide, stone, and concrete. For softer materials, eg, Al–Si alloys, aluminum- or copper-front surface mirrors, and motor commutators, a high positive rake ($+15^\circ$) geometry is used. Positive rake inserts with polycrystalline diamond tips are among the most commonly used tools for this application. The tips can be resharpened and are available in cartridges.

6.1. Low Pressure Diamond Coatings

In the early 1980s Japanese researchers (158) took earlier Russian findings (159, 160) seriously and developed microwave-assisted chemical vapor deposition of diamond having growth rates reaching nearly 1 μm/h. They used a 2.45-GHz microwave plasma for the production of diamond films. This method has been used since that time to coat carbide tools and for optical applications, as the films grown have good quality. Other advantages of this process are its stability, large deposition area, and no metal contamination of the film.

Around the same time a similar technique was independently developed whereby micrometer sized diamond crystallites were grown (161). What is required in essence for the low pressure diamond synthesis is a source of carbon (typically a hydrocarbon gas), hydrogen, and a temperature above 2000°C to convert molecular hydrogen to its atomic state.

In 1988 a technique for diamond synthesis was announced based on an oxy-acetylene combustion flame with a slightly fuel-rich mixture (162). It uses an oxy-acetylene welding torch yielding high (50–150-μm/h) growth rates. This technique offers a very simple and economical means for diamond synthesis at growth rates one to two orders of magnitude higher than microwave or hot-filament-assisted CVD diamond techniques, which are typically 1–2-μm/h. Because the diamond growth takes place under atmospheric conditions, expensive vacuum chambers and associated equipment are not needed. The flame provides its own environment for diamond growth and the quality of the film is dependent on such process variables as the gas flow rates, gas flow ratios, substrate temperature and its distribution, purity of the gases, distance from the flame to the substrate, etc.

Plasma-jet diamond techniques yield growth rates of about 980 μm/h (163, 164). However, the rate of diamond deposition is still one to two orders of magnitude lower than the HP–HT technology, which is about 10,000 μm/h (165). Diamond deposition rates of ca 1 μm/s have been reported using laser-assisted techniques (166). This rate is comparable to the HP–HT synthesis.

In the microwave-assisted or hot-filament-assisted CVD of diamond, methane and hydrogen gases (CH_4 ca 1–5% and H_2 ca 95–99%) are used. In addition, oxygen is used at times to produce improved diamond coatings. Methane provides the source of carbon. The microwave unit generates plasma in a stainless steel chamber. The microwave energy is coupled by the Symmetric Plasma Coupler to produce a uniform ball of plasma at, or slightly above, the substrate surface. The plasma in the case of microwave and hot tungsten filament at temperatures in excess of 2000°C in the case of hot filament ensures hydrogen in the atomic state. The purpose of atomic hydrogen is to etch away any graphite or nondiamond carbon from the coating. The CVD chamber is kept at a pressure of about 6.7 kPa (50 torr), and the substrate temperature is maintained at 950°C. Typical distance between the substrate and a fixed point of the cavity is ca 20–25 mm. Some of the variables used include the type of substrate material, eg, cutting tool material, substrate preparation, substrate temperature, microwave power, distance between the substrate and a fixed point of the cavity, pressure inside the reactor, composition of the gases, their flow rates and flow ratios, and duration of the test. The deposition rate by this technique using CH_4–H_2 is ca 1 μm/h. Use of hot filament or microwave CVD has enabled the formation of crystalline diamond coatings with relative ease (Fig. 14). Depending on the processing conditions used, octahedral, cubic, or cauliflower-like structures can be formed. Whereas the rough surface topography can reduce the friction between the tool rake face and the sliding chip, it can be the base to which metal may be anchored during cutting of soft material and subsequent metal buildup. In such cases it may be necessary to polish the rake face. This, of course, adds to the cost of the cutting tool.

Figure 15 shows the variation of diamond deposition rates by various activated CVD techniques as well as the HP–HT technique (165). It can be seen that the highest growth rate of activated CVD diamond synthesis is still an order of magnitude lower than the HP–HT technique. However, CVD has the potential to become an alternative for diamond growth in view of the significantly lower cost of activated CVD equipment and lower running and maintenance costs.

The activated CVD diamond techniques can be more attractive in cases where the huge capital investment (several hundred million dollars) required for the HP–HT technology is not available or where the high level of technical knowledge required for HP–HT synthesis is not available. In addition, most wear-resistant applications require diamond coatings only of the order of a few micrometers thick. Such coatings can be deposited directly on the finished product without the need for further finishing if CVD techniques are employed.

The low pressure CVD diamond data from various researchers has been summarized (167). Figure 16**a** is the atomic C–H–O diamond deposition phase diagram for all diamond CVD methods used. Most of the combustion synthesis experiments were conducted along the acetylene line. Most of the plasma and hot filament experiments were conducted using highly diluted mixtures of hydrocarbon and hydrogen, sometimes with additional oxygen. The diamond region is very narow in the hydrogen-rich end of the phase diagram and broadens considerably on the C–O line. This diagram indicates that the low pressure diamond synthesis is feasible only within a well-defined field of the phase diagram.

The effect of the substrate temperature can also be considered (Fig. 16**c**). As the substrate temperature increases, the triangular diamond domain region in the C–H–O equilibrium diagram shrinks to almost a line at the highest temperature.

In order to improve adhesion between diamond and the cutting tool substrate, various approaches are being taken. One reason advanced for the poor bonding in the case of cemented carbide substrates is the presence of Co (168). To overcome this problem, either a low Co cemented carbide grade is chosen or the surface Co is etched away before diamond coating takes place (169). One of the desirable characteristics for good bonding between the substrate and the coating is chemical compatibility and good match in crystal structure and lattice parameters between them. An element that comes closest to meeting these requirements in the case of diamond is silicon. Bonding between polycrystalline diamond films grown on a silicon substrate using the microwave CVD is very strong. It is therefore possible to deposit a thin film of silicon on the tool substrate before coating with diamond. Another factor in good bonding is good thermal expansion match between the coating and the substrate. Diamond with the highest thermal conductivity is difficult to match with most tool materials. However, an interlayer of high conductivity material, in which diamond particles can be cemented may be appropriate. The coating can, therefore, adhere strongly to the substrate without spalling.

Another approach is to coat the cutting tool material with a carbide former, such as titanium or silicon or their respective carbides by CVD and deposit diamond on top of it. The carbide layer may serve as an

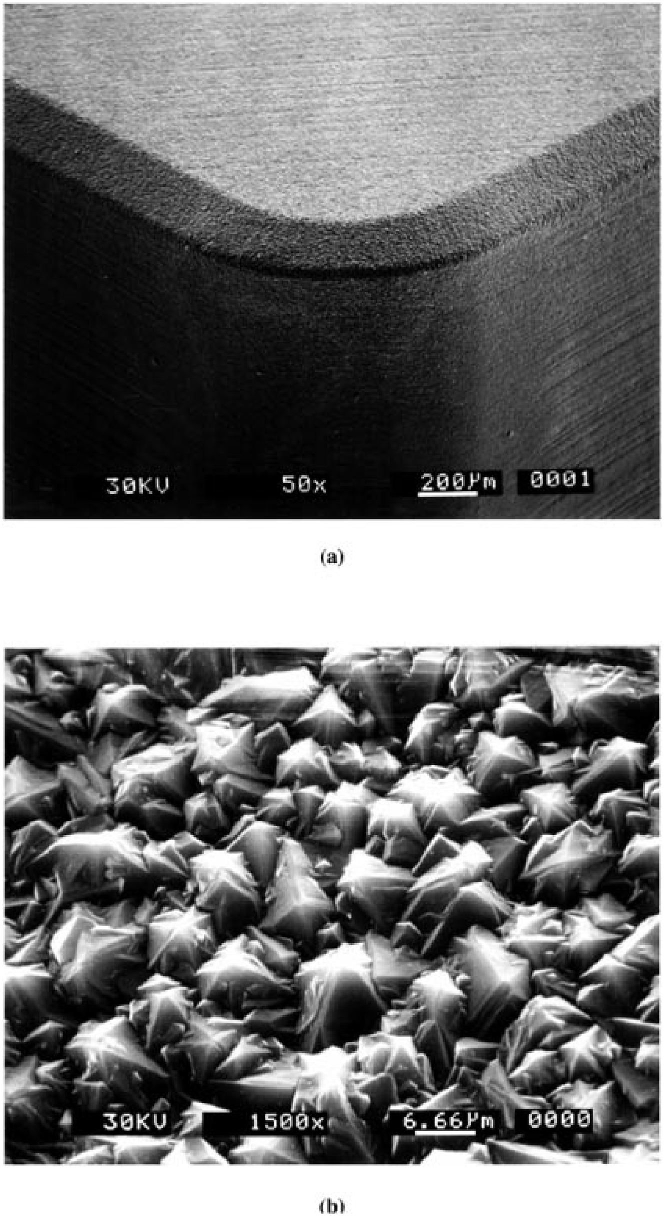

Fig. 14. A scanning electron micrograph of a diamond coating on a silicon nitride cutting tool; (**b**) at higher magnification, the octahedral growth of diamond.

interface between diamond and the cemented carbide, thus promoting good bonding. Yet another method to obtain adherent diamond coatings is laser-induced microwave CVD. By ablating the surface of the substrate with a laser (typically, ArF excimer laser) and coating this surface with diamond by microwave CVD, it is possible to improve the adhesion between the tool and the substrate. Partial success has been achieved in this direction by many of these techniques.

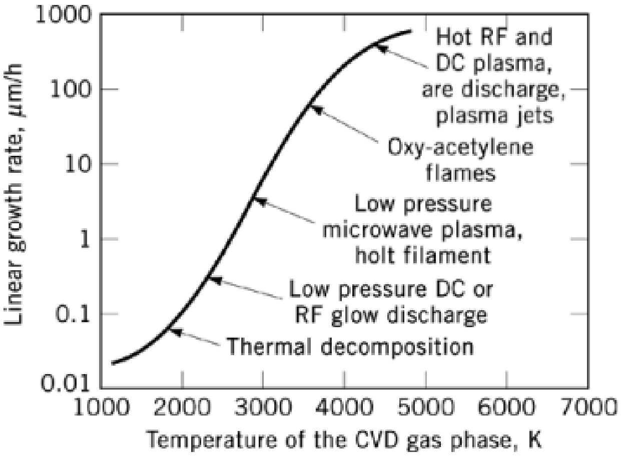

Fig. 15. Variation of diamond deposition rates by various activated CVD techniques as well as the HP–HT technique (165).

Efforts to deposit diamond coatings on cutting tools are concentrated in the United States (chiefly Crystallume, Norton, Kennametal, and General Electric), Japan (Sumitomo Electric, Toshiba Tungaloy, Mitsubishi Metal Corporation, and Asahi Diamond Company), and Europe (Sandvik). Two approaches are generally taken for CVD diamond coatings on cutting tools. One is to grow thick (1–1.5-mm), free-standing polycrystalline diamond slabs cut to size, and braze them onto a cemented carbide substrate. However, the tools have to be finished before use. Most companies are attempting to commercialize this type of CVD diamond tool. As of this writing (ca 1997), such tools are available commercially on a limited basis. This tool type, in concept, is not much different from the polycrystalline diamond made by the HP–HT process.

The other approach is to develop thin (2–5-μm) coatings on cutting tools. This requires use of the CVD diamond process. Microwave CVD, hot filament CVD, plasma torch, and combustion synthesis are some of the techniques used either individually or in combination to deposit diamond coatings on cutting tools. The advantage of this technique is that no subsequent finishing of the tool is required, thus saving considerably on finishing costs. It is this aspect that made hard, refractory thin (1–10-μm) coatings on cemented carbide tools attractive and economical.

Adhesion of the diamond coating to the tool substrate is the main problem challenging researchers. Many difficulties have to be overcome before a good metallurgical bond between diamond coating and the substrate can be developed on cutting tools. Diamond is the most difficult material to bond with most materials because of its unique characteristics. There is therefore a lack of suitable substrate materials having similar properties. To form a good metallurgical bond at the interface in a coated tool, the characteristics of the coating and the substrate should be matched as closely as possible. This includes matching of thermal expansion coefficients, lattice parameters, chemistry, etc. A graded interface having properties of the coating near the substrate closer to the substrate properties and properties near the coating closer to the coating properties must be used in this case.

7. Cubic Boron Nitride (cBN)

Cubic boron nitride (cBN), next only to diamond in hardness (Knoop hardness 46.1 GPa (ca 4700 kgf/mm^2)), was developed in the late 1960s (153–155). It is a remarkable material in that it does not exist in nature and is produced by high temperature–high pressure (HP–HT) synthesis in a process similar to that used to produce diamond from graphite. Hexagonal boron nitride (hBN) is used as the starting material. Alkaline-earth metals and their compounds (instead of Ni in the case of diamond) are found to be the suitable catalyst–solvent for the production of cBN by the HP–HT process. Cubic boron nitride, although not as hard as diamond, is less reactive with ferrous materials like hardened steels, hard chill-cast iron, and nickel-base and Co-base superalloys. It can be used efficiently and economically at higher speed (ca five times), with a higher removal rate (ca five times) than cemented carbide, and with superior accuracy, finish, and surface integrity. Sintered cBN tools are fabricated in the same manner as sintered diamond tools and are available in the same sizes and shapes. Their costs are significantly higher than those of either cemented-carbide or ceramic tools because of higher processing and shaping costs. Like the sintered polycrystalline diamond tools, cBN tools are held on standard tool holders. In order to gain full potential of this material, very rigid precision machine tools with adequate speed and power capabilities are recommended.

Polycrystalline cBN is used extensively for machining of high hardness steels ($\mathrm{Rc} > 45$), nickel-base superalloys, and alloyed cast iron. However, the development of other, less expensive tool materials, chiefly ceramics (SiC whisker-reinforced alumina and SiAlON) for machining of nickel-base alloys is challenging the use of this material. To compete with lower cost advanced tool materials, such as ceramics, coated carbides, newer fabrication technology is under development. For example, instead of fabricating sintered cBN tools on a cemented–carbide base, tools of ca 1.5 mm thickness are fabricated without the cemented carbide support base and the tool faces are ground on either side (170). Such a tool can be used on both sides, which roughly doubles its life. The tool, however, has to be properly supported and clamped during use to prevent premature failure. Polycrystalline cBN tools are very hard and consequently somewhat brittle. Also, cBN tools can be cut into segments of a pie and several tools can be made instead of one round tool.

In order to extend applications of cBN to include machining of medium-hardness steels, modifications of the cBN were introduced. An example is the fabrication of sintered cBN tools by the same HP–HT process, but using binder and second phase (either metallic or nonmetallic) such as TiN or TiC to increase toughness (171). In regard to phase distribution, cBN tools resemble cemented-carbide or alumina–TiC ceramic tools, but are tougher and have greater chemical stability.

The two predominant wear modes of cBN tools are DCL notching and microchipping. Polycrystalline cBN tools exhibit flank wear where alumina ceramic tools fail catastrophically. These tools have been used successfully for heavy interrupted cutting and for milling white cast iron and hardened steels. Negative lands and honed cutting edges were used. Like diamond, cBN is thermally unstable at elevated temperatures. The reaction products, however, when machining materials like steel- or nickel-base alloys, are generally not damaging to the process. cBN tools are not recommended for very low or very high speed cutting applications. Nevertheless, these tools are capable of very high removal rates when used with machine tools of adequate power and stiffness. In fact, they perform better with heavier cuts than with lighter cuts. Because they are inherently brittle, cBN tools are used with a negative rake (-5°) geometry. Suitable edge preparation, consisting of honing a small radius or a small negative land on the rake face, is also recommended.

Diamond and cBN tools provide significantly higher performance capability, and demands are being placed on the machine tools and manufacturing practice in order to take full advantage of the potential of these materials. Being extremely hard but brittle, the rigid machine tools must be used with gentle entry and exit of the cut in order to prevent microchipping by cleavage. High precision machine tools offer the advantage of producing high finish and accuracy. Use of machine tools with higher power and rigidity enables higher removal rates. A more recent application of cBN is the finish machining of hardened steels, such as in bearing races. Typically these materials are machined in the annealed condition to remove much of the unwanted material and subsequently heat-treated to the desired hardness and ground to the required size, accuracy, and finish. This can be a time-consuming and expensive operation. Using cBN tools, the bearing steels are obtained in their final hardness condition and are machined to the required size, accuracy, and finish without the need for subsequent grinding on rigid, high precision machine tools.

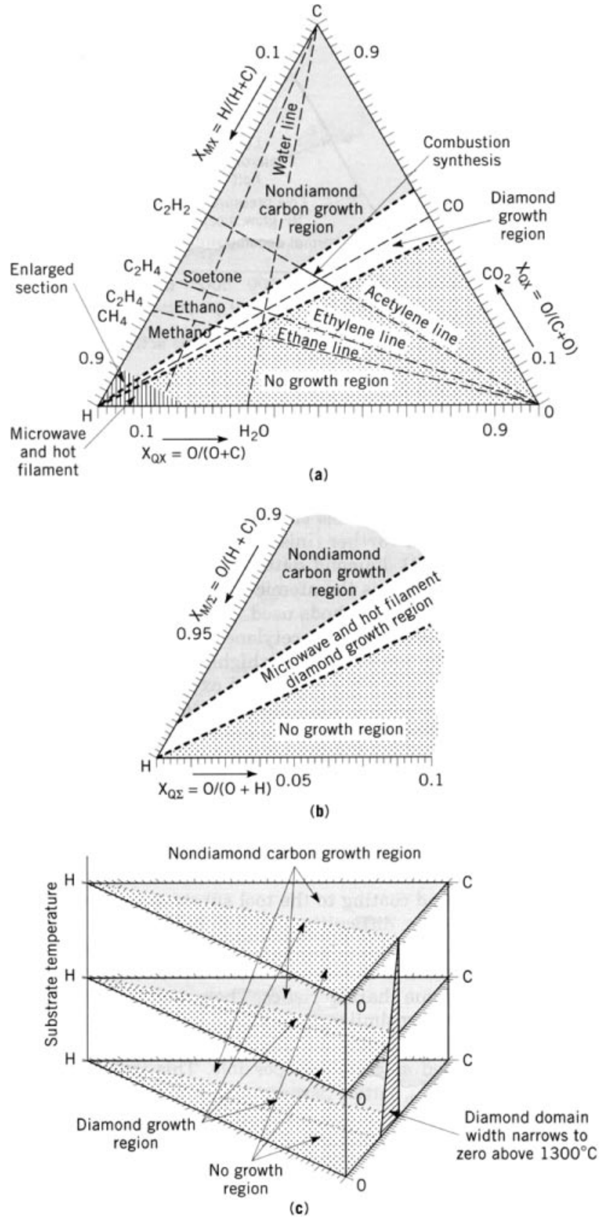

Fig. 16. (**a**) Atomic C–H–O diamond deposition phase diagram for various CVD diamond methods used showing a definite diamond growth region (adapted from Ref. 167); (**b**) enlarged hydrogen-rich region of the phase diagram shown in (**a**) where most of the plasma and hot filament experiments were conducted (adapted from Ref. 167); (**c**) effect of the substrate temperature on the diamond domain in the C–H–O diamond deposition phase diagram (adapted from Ref. 167).

8. Economic Aspects

Machining costs (labor and overhead) in the United States have an estimated value of $>\$300 \times 10^9/\mathrm{yr}$. The cost of labor and overhead for machining is based on the estimated number of total metal-cutting machine tools in various metal-cutting industries (1). This value does not take into account the cost of raw stock (work material), cutting tools, and many other support facilities. An estimated breakdown of cutting tool costs is given in Table 13 (172). Because of the competitive nature of these industries, most prefer to keep cost information proprietary.

Table 13. Estimated Breakdown of Cutting Tool Costs

Material	Cost, $
high carbon, low alloy, HSS	$(1 - 125) \times 10^9$
cemented carbides	$750 \times 10^{6\,a}$
ceramics	25×10^6
diamond, cBN	$(25 - 50) \times 10^6$

[a] About half is for coated grades.

In 1981, the value of disposable metal-cutting tools shipped to various U.S. manufacturing plants was estimated to be $\$2.13 \times 10^9$, only 1% of the total estimated U.S. manufacturing costs. Thus, the cost of cutting tools per se is only a small fraction of the total costs, although the tooling costs may be significant in a large manufacturing facility. The costs associated with the use of cutting fluids is estimated to be about 16% of the manufacturing costs (173–175).

High speed steels (HSS) and cemented carbides were the most extensively used tool materials in 1996, accounting for ca $\$2 \times 10^9$ in sales. From $\$1 - 1.25 \times 10^9$ were for HSS and the remaining portion for carbides. The market for ceramics is ca $\$25 \times 10^6$. Although uncoated tools of the HSS and cemented carbide materials are still used, the trend is toward more extensive (ca 60% in the United States and western Europe) use of coated tools. For HSS coated tools this percentage has not quite reached 50%. This is partly because HSS tools are relatively inexpensive and hence some of the customers have not fully appreciated the tangible benefits of coatings, especially when used in small batches as in small job shops, and also partly because many HSS tools are reground, ie, they have to be recoated for use as coated tools. Diamond and cubic boron nitride (cBN) are used for special applications where despite high cost, use is justified because of high hardness. The market for polycrystalline diamond and cBN is in the range of $\$25 - 50 \times 10^6$. Most of the cost information is kept confidential by individual companies and not disclosed; the costs given here are, therefore, estimates based on indirect information. Cast-cobalt alloys are presently phased out because of the high cost of raw materials (Co, Cr, and W), and because of safety problems encountered in the handling of Co and the increasing availability of alternative materials with superior performance at reduced cost. New ceramics will have significant impact on future manufacturing productivity, especially as improved fracture toughness and strength and hence the reliability of these materials occur.

9. Health and Safety Factors

Threshold limit values for the components of cemented carbides and tool steels are given in Table 14 (176). There is generally no fire or explosion hazard involved with tool steels, cemented carbides, or other tool materials. Fires can be handled as metal fires, eg, with Type D fire extinguishers. Most constituents of tool materials do not polymerize.

During machining operations, eye protection is recommended; during grinding operations, NIOSH-approved respirators for metal fumes and dust are recommended (177). Fine powder of Co is known to cause

Table 14. Threshold Limit Values (TLV)

Constituent	TLV, mg/m^3
tungsten carbide	5
titanium carbide	na
tantalum carbide	5
chromium carbide	0.5
cobalt	0.1
nickel	1
iron	na
tungsten	5
copper	1

dermatitis and pulmonary disorders in humans. Most manufacturers supply safety information with their products (178). These should be followed strictly for the welfare of the personnel on the shop floor.

Safety is of particular concern in metal-cutting and metal-forming operations (178). Precautions should be taken to ensure protection of personnel and equipment from potential flying fragments and sharp edges as well as the large volumes of chips produced. Safety devices and protective shields or screens must be installed on metal-cutting machines. Chips should be handled with some mechanical device, never by hand. In automated machining, chip handling by effective chip groove geometry should be practiced. Tool overhang must be as short as possible to avoid instances of deflection, resulting in breakage or chatter. Noise caused by chatter or vibration can be highly objectionable to personnel nearby, in addition to the operator. Corrective measures should be taken wherever chatter prevails. These include change in the cutting conditions, modification of the tool–work support system to increase its stiffness, or operating at conditions below or above the natural frequencies of parts of the machine tool that cause chatter.

Some cutting fluids, eg, oils, may present a fire hazard. Some work materials, eg, magnesium, aluminum, titanium (under certain conditions), and uranium, in finely divided form, also present fire hazards. Very small metal chips or dust may ignite.

Adequate ventilation of grinding operations should be established to comply with existing government regulations, and management should remain alert to the possibility of symptoms, even in grinders working within established government standards.

The high temperatures generated in machining, especially at high cutting speeds, necessitate the use of a refractory cutting tool that can withstand the high temperature and provide long tool life. Cutting fluids are needed to absorb the high heat, cool the cutting tool at higher speeds, lubricate at low speeds and high loads, increase the tool life, improve the surface finish, reduce the cutting forces and power consumption, etc. The physiological effects of cutting fluids on the operator must be considered. Toxic vapors, unpleasant odors, smoke fumes, skin irritations (dermatitis), or effects from bacteria cultures from the cutting fluid are all factors. Consumption of cutting fluids has been reduced drastically by using mist lubrication. However, mist in the industrial environment can have a serious respiratory effect on the operator. Consequently, standards are being set to minimize this effect. Many industrialized nations, including Germany, have made commitment to provide a safer working environment and the United States has no option but to provide a similar manufacturing environment.

10. Future Outlook

The raw materials used in the cutting tools are currently (ca 1997) made by melting and subsequent size reduction by milling to fine size. Many defects such as microcracks, voids, etc, are generated during solidification

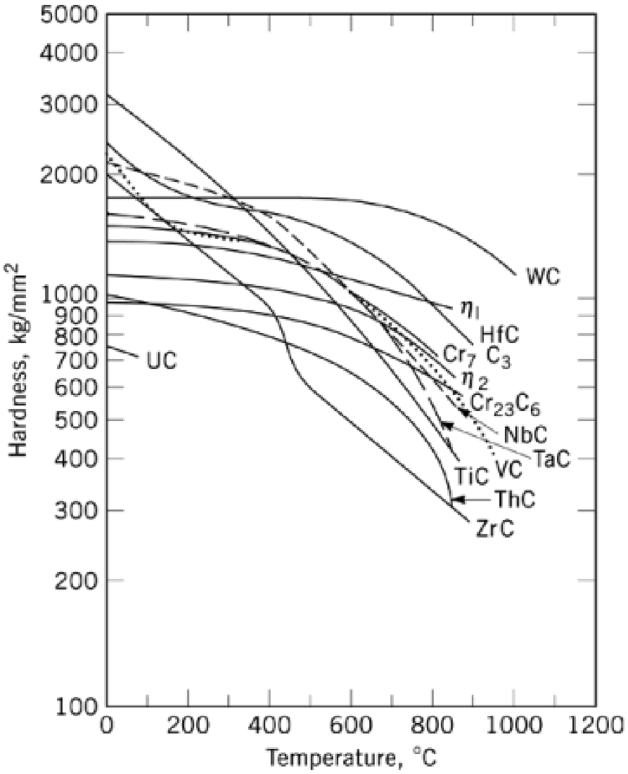

Fig. 17. Variation of hardness of single crystals of various monocarbides with temperature (44).

and the subsequent size-reduction process. The comminution process limits the size of the crystallites to ca 1-μm. Newer technologies are being developed based on chemical routes, such as the sol–gel technique for the production of ultrafine materials in nanocrystalline size. These materials are also relatively free from defects and hence tool materials based on nanocrystalline materials may increasingly be used, especially for lower speed, roughing, or intermittent cutting operations or operations where edge strength is an essential requirement, as in milling.

In many of the thin coatings on cemented carbide, either single or multiple coatings of single phase materials, such as TiC, are used. It would appear that extending the use of solid solutions of multicarbides of W, Ti, and Ta or Nb for coatings may further enhance the performance of the coated carbides. It would not be difficult to accomplish this either by CVD or PVD techniques.

Even though TiC is much harder than WC at room temperature (3200 kg/mm^2 for TiC, vs 1800 kg/mm^2 for WC), at higher temperatures, TiC oxidizes and loses its hardness rapidly. Figure 17 is a plot of the variation of hardness of single crystals of various monocarbides with temperature (44). No similar data is available for multicarbides or other refractory hard materials, such as nitrides, borides, oxides, or any combination of them.

It is known that fracture toughness of materials can be increased significantly by transformation toughening as in the case of ZrO_2, by applying crack deflection methods involving fine dispersion of second-phase particles, as in the case of Al_2O_3 + TiC, and whisker pullout in as in the case of SiC-whisker-reinforced alumina. This approach is expected to find greater use for a range of tool materials where toughness is an important consideration, eg, in ceramic tools.

Metal cutting research in the 1950s (179) clearly showed that significant reductions in forces can result using an increase in the rake angle ($\leq 45^\circ$). Consequently, the energy requirements and heat generated, etc, would also be reduced. For the most part as of 1996 -5° rake and 5° clearance inserts were used. This is true partly owing to the application of the same insert for a range of work-materials and partly owing to concerns that high positive geometry might render the tool weak, especially when machining high strength materials.

Use of higher rake angles (ca 45°) would permit higher cutting speed for the same tool life or longer tool life for the same cutting speed, improved surface finish, lower cutting forces resulting in lower cutting energy and power requirements or higher removal rates, lower thrust forces and consequently lower deflections, and reduced tool wear owing to lower interface temperature. However, strength of the cutting edge is rather critical owing to low included angle of the tool. Submicrometer-grain cemented carbide material or submicrometer-grain TiC or TiN in HSS steels are recommended for such applications, especially at low cutting speeds. All these approaches lead to increased efficiency and longer tool life.

The implementation of coated tool technology on the shop floor is proceeding at a significant pace. New coating combinations, tailored substrates, CVD/PVD technologies for different types of engineered coatings for different workmaterials, cutting conditions, machining operations are being addressed effectively. This effort is expected to continue. Multiple nanolayer coatings which have more recently been developed have improved hardness, strength, and chemical stability. The practically unlimited choice of coating combinations, ie, alternate hard and tough materials, or alternate hard material and solid lubricant, etc, should lead to numerous multiple nanolayer coating applications. Nanolayer coatings may be ideal for hard refractory coatings which have difficulty in bonding with the substrate or other coatings.

SiC whisker-reinforced alumina is a major advance in tool material development, as it provides a means to increase the fracture toughness of the material via the composite material approach. It is entirely possible that in the next century many new whiskers of refractory, hard materials will be made available economically for application to cutting tools. One may even consider SiC whisker-reinforced alumina as a model material on the basis of which many new tool materials may be developed. Tool material for high speed machining of titanium alloys may evolve from this concept as most tool materials are very reactive with respect to titanium. Some of the intermetallics may be candidate materials for this application.

Another interesting concept, one used in the development of superalloys, is the strengthening of the matrix by orderly precipitation of the second-phase materials and strengthening by dispersoids, such as Al_2O_3. Strengthening by dispersoid is to some extent already practiced, as in the cases of TiC in Al_2O_3, and TiN in cemented TiC. However, orderly precipitation by solution treatment and aging can be an attractive approach. For example, one can consider extensive dispersion of fine precipitates of TiC, TiN, or TiB_2 in a titanium matrix; ZrB_2 in a zirconium matrix; or B_4C in a boron matrix; or hot-pressing of B_4C in titanium matrix.

Cemented carbides are fairly expensive owing to the use of hard, refractory materials. This is expected to become even more the case as some of the strategic materials used in these tools become more expensive or newer but more expensive materials such as HfC or HfN come into more common use. It may be economical, therefore, to use these materials at or near the cutting edge instead of as the whole insert. The development of tools of TiC (40–55%) or TiN (30–60%) in a steel matrix on a steel core using powder metallurgy technology suggests a similar approach for cemented carbides as the need arises.

BIBLIOGRAPHY

“Tool Materials for Machining” in ECT 1st ed., Suppl. 2, pp. 873–882, by Roland B. Fischer, Battelle Memorial Institute; “Tool Materials for Machining” in ECT 2nd ed., Vol. 20, pp. 566–578, by Roland B. Fischer, The Dow Chemical Company; “Tool Materials” in ECT 3rd ed., Vol. 23, pp. 273–309, by R. Komanduri and J. D. Desai, General Electric Co.

Cited Publications

1. *Machinability Data Handbook*, 3rd ed., Vols. **1** and **2**, Machinability Data Center, Metcut Research Associates, Cincinnati, Ohio, 1980.
2. E. Dow Whitney, ed., *Ceramic Cutting Tools*, Noyes Publications, Park Ridge, N.J., 1994.
3. R. Komanduri, *ASME Appl. Mechan. Rev.* **46**(3), 80–132 (Mar. 1993).
4. E. D. Doyle, personal communication, 1981.
5. V. A. Tipnis, in M. B. Peterson and W. O. Winer, eds., *Wear Control Handbook*, ASME, New York, 1980, 891–930.
6. U.S. Pat. 4,286,905 (Sept. 1, 1981), S. K. Samanta (to Ford Motor Co.).
7. H. G. Swinehart, ed., *Cutting Tool Materials Selection*, American Society of Tool and Manufacturing Engineers (now the Society of Manufacturing Engineers (SME)), Dearborn, Mich., 1968.
8. R. Komanduri, "Cutting Tool Materials" in M. B. Bever, ed., *Encyclopedia of Materials Science and Engineering*, Pergamon Press, Oxford, U.K., 1983 General Electric TIS report no. 82CRD176, Schenectady, N.Y., June 1982.
9. R. Komanduri and J. D. Desai, *Tool Materials for Machining*, General Electric TIS report no. 82CRD220, Schenectady, N.Y., Aug. 1982.
10. *Machinability Data Handbook*, 3rd ed., Vols. **1 and 2**, Machinability Data Center, Metcut Research Associates, Cincinnati, Ohio, 1980.
11. *Manuf. Eng.*, **49** (Oct. 1977).
12. K. J. A. Brooks, *World Directory and Handbook of Hard Metals*, 5th ed., An Engineer's Digest Ltd., U.K., 1992.
13. R. L. Hatschek, *Am. Machinist*, 165–176 (May 1981).
14. *Properties and Selection of Tool Materials*, American Society for Metals, Metals Park, Ohio, 1975.
15. Properties and Selection: Stainless Steels, Tool Materials and Special-Purpose Metals, Vol. **3** of *Metals Handbook*, 9th ed., American Society for Metals, Metals Park, Ohio, 1979, 421–488.
16. J. T. Berry, *"Recent Developments in the Processing of High-Speed Steels,"* Climax Molybdenum Co., Greenwich, Conn., 1970.
17. *Properties and Selection of Tool Materials*, American Society for Metals, Metals Park, Ohio, 1962.
18. G. A. Roberts, J. C. Hamaker, and A. R. Johnson, *Tool Steels*, American Society for Metals, Metals Park, Ohio, 1962.
19. A. M. Bayer and B. A. Beecherer, in *ASM Handbook*, 9th ed., Vol. **16**, *Machining*, American Society for Metals, Metals Park, Ohio, 1989, 51–59.
20. E. J. Dulis and T. A. Neumeyer, *Materials for Metal Cutting*, ISI Publication P216, The Iron and Steel Institute, U.K., 1970, 112–118.
21. M. G. H. Wells and L. W. Lherbier, eds., *Processing and Properties of High Speed Tool Steels*, Metallurgical Society of AIME, Warrendale, Pa., 1980.
22. R. W. Stevenson, in ASM Handbook, 9th ed., Vol. **7**, *Powder Metallurgy*, 1984, 784–793.
23. K. E. Pinnow and W. Stasko, in Ref. 19, 60–68.
24. R. Komanduri and M. C. Shaw, *Proceedings of the Third North American Metal Working Research Conference (NAMRC-III)*, May 5–7, 1975, Carnegie Press, Pittsburgh, Pa., 1975.
25. R. Wilson, *Metallurgy and Heat Treatment of Tool Steels*, McGraw-Hill Book Co., Inc., New York, 1975.
26. W. T. Haswell, W. Stasko, and F. R. Dax in Ref. 20, 147–158.
27. S. J. Whalen, *Vapor Deposition of Titanium Carbide*, ASTME Technical Paper No. 690, American Society of Tool and Manufacturing Engineers (now Society of Manufacturing Engineers (SMF)), Dearborn, Mich., 1965.
28. U.S. Pats. 2,962,388 and 2,962,399 (Nov. 29, 1960), W. Ruppert and G. Schwedler (to Metallgesellschaft Aktinengesellschaft).
29. M. Epner and E. Gregory, *Cermets*, Reinhold Publishing Co., New York, 1960, 146–149.
30. J. L. Ellis, *Tool Eng.* **38**, 103–105 (1957).
31. S. E. Tarkan and M. K. Mal, *Metal Progr.* **105**, 99–102 (1974).
32. *Modern Metal Cutting—A Practical Handbook*, Sandvik Coromant, Sweden, 1994, pp. III-41 to III-44.
33. Cobalt Monograph, prep. by Battelle Memorial Institute, Columbus, Ohio for Centre d'Information du Cobalt, Brussels, Belgium, 1960.
34. *Technical information brochures*, Stellite Division of Cabot Corp., Kokomo, Ind.
35. H. Moissan, *The Electrical Furnace*, tr. V. Lenher, Chemical Publishing Co., 1904.
36. U.S. Pat. 1,549,615 (Aug. 11, 1925), K. Schroter (to GE).

37. U.S. Pat. 1,721,416 (July 16, 1929), K. Schroter (to GE).
38. W. J. Loach, MR 71-901 in Ref. 67.
39. W. Dawihl, *Handbook of Hard Metals*, English transl., Her Majesty's Stationery Office, London, 1955.
40. P. Schwarzkopf and R. Keiffer, *Cemented Carbides*, The Macmillan Company, New York, 1960, 116–120.
41. H. E. Exner, *Int. Met. Rev.*, (243), 149 (1979).
42. R. W. Stevenson, "Cemented Carbides," in Ref. 22, 773–783.
43. A. T. Santhanam, P. Tierney, and J. L. Hunt, in A. T. Santhanam, P. Tierney, and J. L. Hunt, ASM Handbook, 10th ed., Vol. **2**, *Properties and Selection*, *American Society for Metals*, Metals Park, Ohio, 1990, 950–977.
44. L. E. Toth, *Transition Metal Carbides and Nitrides*, Academic Press, New York, 1971.
45. U.S. Pat. 2,015,536 (Sept. 24, 1935), K. Schroter (to GE).
46. U.S. Pat. 1,925,910 (Sept. 5, 1933), P. Schwarzkopf and I. Hirschel (to GE).
47. U.S. Pat. 2,113,353 (Apr. 5, 1938), P. M. McKenna (to Kennametal).
48. U.S. Pat. 2,113,354 (Apr. 5, 1938), P. M. McKenna (to Kennametal).
49. U.S. Pat. 2,113,355 (Apr. 5, 1938), P. M. McKenna (to Kennametal).
50. U.S. Pat. 2,113,356 (Apr. 5, 1938), P. M. McKenna (to Kennametal).
51. U.S. Pat. 2,124,509 (July 19, 1938), P. M. McKenna (to Kennametal).
52. J. Gurland and P. Bardzil, *Trans. AIME*, **311** (Feb. 1955).
53. M. J. Kuderko, (EM 71–937) in Ref. 67.
54. O. Kasukawa, (EM 71-938) in Ref. 67.
55. J. E. Mayer, D. Moskowicz, and M. Humenik, *Materials for Metal Cutting*, ISI Special Report No. P126, The Iron and Steel Institute, U.K., 1970, 143–151.
56. U.S. Pat. 2,967,349 (Jan. 10, 1961), M. Humenik and D. Moskowicz (to Ford Motor Co.).
57. D. Moskowitz and M. Humenik, Jr., *Proceedings of the 1975 International Powder Metallurgy Conference*, New York.
58. W. W. Gruss, in Ref. 19, 90–97.
59. C. G. Goetzel, in Ref. 22, 798–815.
60. M. Humenik and N. Parikh, *J. Am. Cer. Soc.* **39**, 60 (1956).
61. H. Doi, in H. Doi, *Science of Hard Materials*, E. A. Almond, C. A. Brookes, and R. Warren, eds., 1984, 489–523.
62. U.S. Pat. 4,902,395 (Feb. 20, 1990), H. Yoshimura (to Mitsubishi).
63. C. W. Beeghly, R. V. Godse, and F. B. Battaglia, SME Technical Paper No. MR 90-247, *Proceedings of the 3rd Conference on Adv. Machining Technology*, Chicago, Ill., Sept. 4–6, 1990, p. 247.
64. H. Tanaka, in H. Tanaka, *Cutting Tool Materials: Proceedings of the ASM International Conference*, Ft. Mitchell, Ky., ASM International, Metals Park, Ohio, Sept. 1980, 349–361.
65. R. Komanduri, *Carbide Tool J.* **18**(4), 33–38 (July/Aug. 1986).
66. R. Komanduri, *Advances in Hard Materials Tool Technology*, Carnegie Press, Pittsburgh, Pa., 1976.
67. *Proceedings from the 1st International Cemented Carbide Conference*, Chicago, Ill., Feb. 1–3, 1971, Vols. **1 and 2**, Cemented Carbide Producers Association and the Society of Manufacturing Engineers, Dearborn, Mich.
68. N. P. Suh, *Wear* **62**, 1–20 (1980).
69. N. A. Horlin, *Prod. Eng. (London)*, **153** (1971).
70. A. T. Santhanam and D. T. Quinto, in A. T. Santhanam and D. T. Quinto, ASM Handbook, American Society for Metals, Metals Park, Ohio, 10th ed., Vol. **5**, *Surface Engineering*, 1994, 900–908.
71. H. G. Prengel, W. R. Pfouts, and A. T. Santhanam, *SME Manuf. Eng.*, 82–88 (July 1996).
72. G. Schuhmacher, *"TiC-Coated Hard Metal,"* MR71-930 in Ref. 67.
73. U.S. Pat. 3,977,061 (Aug. 31, 1976), J. N. Lindstrom and F. J. O. E. Ohlsson (to Sandvik).
74. U.S. Pat. 3,616,506 (Nov. 2, 1971), C. S. G. Ekemar (to Sandvik).
75. U.S. Pat. 3,836,392 (Sept. 17, 1973), B. Lux, R. Funk, H. Schachner, and C. Triquet (to Sandvik AB).
76. W. Scott Buist, MR71-931 in Ref. 67.
77. U.S. Pat. 3,642,522 (Feb. 15, 1972), H. Gass and H. E. Hintermann (to Laboratoire Suisse De Recherches Horlogeres).
78. Y. Ohtsu, FC71-932 in Ref. 67.
79. U.S. Pat. 3,736,107 (May 29, 1973), T. E. Hale (to Carboloy).
80. M. Lee, R. H. Richman, and J. Stanislao, FC 71-928 in Ref. 67.
81. U.S. Pat. 3,717,496 (Feb. 20, 1973), R. Kieffer.
82. U.S. Pat. 3,744,979 (July 10, 1973), H. S. Kalish (to Adamas Carbide).

83. B. M. Kramer, *Thin Solid Films* **108**, 117–125 (1983).
84. U.S. Pat. 4,101,703 (July 18, 1978), W. Schintlmeister (to Schwarzkopf Development Corp.).
85. W. Schintlmeister, W. Wallgram, J. Kanz, and K. Gigl, *Wear* **100**, 153–169 (1984).
86. Dreyer and co-workers, *Development and Tool Life Behavior of Super-Wear Resistant Multilayer Coatings on Hard Metals*, Book No. 278, Metals Society (London), 1982, 112–117.
87. D. T. Quinto, A. T. Santhanam, and P. C. Jindal, *Int. J. Refrac. Metals Hard Mater.* **8**(2), 95–101 (June 1989).
88. U.S. Pat. 4,463,062 (July 31, 1984), T. E. Hale (to Carboloy).
89. B. J. Nemeth, A. T. Santhanam, and G. P. Grab in H. M. Ortner, ed., *Proceedings of the 10th Plansee Seminar on Trends in Refractory Metals and Special Materials and Their Technology*, Metallwerk Plansee, Ruette, Austria, 1981, 613–627.
90. U.S. Pat. 4,610,931 (Sept. 9, 1986), B. J. Nemeth (to Kennametal).
91. A. T. Santhanam, G. P. Grab, G. A. Rolka, and P. Tierney, in V. K. Sarin, ed., *Proceedings of the High Productivity Machining*, American Society for Metals, Metals Park, Ohio, 1985, 105–112.
92. U.S. Pat. 4,828,612 (May 9, 1989), W. C. Yohe (to Carboloy).
93. U.S. Pat. 4,830,930 (May 16, 1989), Y. Taniguchi, K. Sasaki, M. Ueki, and K. Kobori (to Toshiba Tungaloy).
94. U.S. Pat. 5,372,873 (Dec. 13, 1994), H. Yoshimura, T. Tanaka, A. Osada, and T. Sudo (to Mitsubishi Materials).
95. S. Ramalingam, in Ref. 5, 385–411.
96. R. Horsfall and R. Fontana, *Cutting Tool Eng.*, 37–42 (Feb. 1993).
97. J. von Stebul and co-workers, *Surf. Coating Technol.* **68/69**, 762–769 (1994).
98. *Catalog*, Balzers Tool Coatings, Inc., North Tonawanda, N.Y., 1996.
99. U.S. Pat. 5,503,913 (Apr. 2, 1996), U. Konig and R. Tabersky (to Widia GmbH).
100. W. J. Clegg, K. Kendall, N. McN. Alford, T. W. Burton, and J. D. Birchall, *Nature (London)* **347**, 455–457 (1991).
101. P. Anderson, *Scripta Metall. Mater.* **27**, 687–692 (1992).
102. M. C. Shaw, D. B. Marshall, M. S. Dadkhah, and A. G. Evans, *Acta Metall. Mater.* **41**(11), 3311–3322 (1993).
103. C. A. Folsom, F. W. Zok, and F. F. Lange, *J. Am. Ceramic Soc.* **77**(3), 689–696 (1994).
104. U.S. Pat. 4,804,583 (Feb. 14, 1989), T. D. Moustakas (to Exxon R&E Co.).
105. A. T. Alpas, J. D. Embury, D. A. Hardwick, and R. W. Springer, *J. Mater. Sci.* **25**, 1603–1609 (1990).
106. Y. Ding, Z. Farhat, D. O. Northwood, and A. T. Alpas, *Surf. Coatings Technol.* **62**, 448 (1993).
107. Y. Ding, Z. Farhat, D. O. Northwood, and A. T. Alpas, *Surf. Coatings Technol.* **68/69**, 459–467 (1994).
108. U.S. Pat. 4,643,951 (Feb. 17, 1987), J. E. Keem and J. D. Flasck (to Ovonic Synthetic Materials).
109. U.S. Pat. 5,268,216 (Dec. 7, 1993), J. E. Keem and B. M. Kramer (to Ovonic Synthetic Materials).
110. M. R. Hilton and co-workers, *Surf. Coating Technol.* **53**, 13–23 (1992).
111. G. Jayaram, L. D. Marks, and M. R. Hilton, *Surf. Coating Technol.* **76–77**, 393–399 (1995).
112. E. Santner, D. Klaffke, and G. Meier zu Kocher, *Wear* **190**, 204–211 (1995).
113. R. Hubler and co-workers, *Surf. Coating Technol.* **60**, 561–565 (1993).
114. R. Komanduri and J. Larsen-Basse, *ASME Mechan. Eng.* **111**, 74–79 (Jan. 1989).
115. A. G. King and W. M. Wheildon, *Ceramics in Machining Processes*, Academic Press, Inc., New York, 1966.
116. F. W. Wilson, ed., *Machining with Carbides and Oxides*, McGraw-Hill Book Co., Inc., 1962.
117. E. D. Whitney, SAE Technical Paper No. 810319, Society of Automotive Engineers, *International Congress and Exposition*, Feb. 23–27, 1981.
118. R. Komanduri and S. K. Samanta, in Ref. 15, 98–104.
119. R. Komanduri, *Int. J. Refract. Metals Hard Mater.* **8**(2), 125–132 (June 1989).
120. U.S. Pat. 2,873,198 (Feb. 10, 1959), E. W. Goliber (to GE/Carbaloy).
121. E. W. Golliber, in E. W. Golliber, E. W. Golliber, eds., in *Proceedings of the 66th Annual Meeting of the American Ceramic Society (ACS)*, Westerville, Ohio, 1960.
122. J. Binns, *ASTME (now SME) Paper No. 633*, Society of Manufacturing Engineers, Dearborn, Mich., 1963, 1–15.
123. K. H. Jack, *Metals Technol.* **9**, 297–301 (July 1982).
124. Y. Oyama and O. Kamigaito, *Jpn. J. App. Phys.* **10**, 1637 (1971).
125. K. H. Jack and W. I. Wilson, *Nature (London) Phys. Sci.* **238**, 28–29 (1972).
126. U.S. Pat. 3,992,166 (Nov. 9, 1976), K. H. Jack and W. I. Wilson (to Joseph Lucas Industries Ltd.).
127. U.S. Pat. 3,991,148 (Nov. 9, 1976), R. J. Lumby, R. R. Wills, and R. F. Horsley (to Joseph Lucas Industries, Ltd.).

128. J. G. Baldoni and S. T. Buljan, in E. Dow Whitney, ed., *Ceramic Cutting Tools*, Noyes Publications, Park Ridge, N.J., 1994.
129. S. K. Bhattacharyya and A. Jawaid, *Int. J. Prod. Res.* **19**(5), 589–594 (1981).
130. S. K. Bhattacharyya, A. Jawaid, and J. Wallbank, in S. K. Bhattacharyya, A. Jawaid, and J. Wallbank, *Proceedings of the ASME High-Speed Machining Symposium*, New Orleans, La., PED Vol. **12**, American Society of Mechanical Engineers, 245–262.
131. U.S. Pat. 4,227,842 (Oct. 4, 1980), S. K. Samanta, S. Subramanian, and A. Ezis (to Ford Motor Co.).
132. U.S. Pat. 4,401,617 (1983), A. Ezia, S. K. Samanta, and K. Subramanian (to Ford Motor Co.).
133. U.S. Pat. 4,401,238 (Feb. 28, 1984), A. Ezia, S. K. Samanta, and K. Subramanian (to Ford Motor Co.).
134. S. K. Samanta and K. Subramanian, *Proceedings of the North American Manufacturing Research Conference (NAMRC)*, University of California, Berkeley, Calif., 1985.
135. *Engineering*, 1009 (Sept. 1980).
136. V. K. Sarin and S. T. Buljan, in V. K. Sarin and S. T. Buljan, *High Productiv. Machin. Proc. Int. Conf. High Productivity Machining*, New Orleans, La., May 7–9, 1985, ASM International, Metals Park, Ohio, 113–120.
137. U.S. Pat. 4,440,547 (Apr. 1984), V. K. Sarin and S. J. Buljan (to GTE).
138. U.S. Pat. 4,406,668 (Sept. 27, 1983), V. K. Sarin and S. J. Buljan (to GTE).
139. U.S. Pat. 4,406,669 (Sept. 27, 1983), V. K. Sarin and S. J. Buljan (to GTE).
140. U.S. Pat. 4,406,670 (Sept. 27, 1983), V. K. Sarin and S. J. Buljan (to GTE).
141. U.S. Pat. 4,543,345 (Sept. 24, 1985), G. C. Wei.
142. K. H. Smith, *Carbide Tool J.* **18**(5), 8–10 (Sept.–Oct. 1986).
143. U.S. Pat. 4,543,345 (Sept. 24, 1985), G. C. Wei (to Oakridge National Laboratories, DOE).
144. U.S. Pat. 4,749,667 (June 7, 1988), C. Jun (to Carboloy).
145. S. T. Buljan, J. G. Baldoni, and M. L. Juckabee, *Ceram. Bull.* **66**(2), 347–352 (1987).
146. P. K. Mehrotra, M.P.R., 506–510 (July–Aug. 1987).
147. E. R. Billman, P. K. Mehrotra, L. F. Shuster, and C. W. Deeghly, *Ceram. Bull.* **67**(6), 1016–1019 (1988).
148. S. A. Buljan and S. F. Wayne, *Adv. Ceram. Mater.* **2**(4), 813–816 (1987).
149. J. R. Baldoni and S. T. Buljan, *Ceram. Bull.* **67**(2), 381–387 (1988).
150. J. G. Lee and I. B. Cutler, *Ceram. Bull.* **54**(2) (1975).
151. C. K. Jun and K. H. Smith, in Ref. 3, 86–111.
152. C. F. Lewis, *Mater. Eng.*, 31–35 (July 1986).
153. F. P. Bundy, H. T. Hall, H. M. Strong, and R. H. Wentorf, Jr., *Nature* **176**, 51 (1955).
154. F. P. Bundy, *Sci. Am.* **231**, 62 (Aug. 1974).
155. R. H. Wentorf, Jr., R. C. DeVries, and F. P. Bundy, *Science* **208**, 873 (May 23, 1980).
156. L. E. Hibbs, Jr. and R. H. Wentorf, Jr., *High Temp. High Pressure* **6**, 409 (1974).
157. U.S. Pat. 4,224,380 (Sept. 23, 1980), H. P. Bovenkerk and P. D. Gigl (to GE).
158. S. Matsumoto, Y. Sato, M. Tsutsumi, and N. Setaka, *J. Mater. Sci.* **17**, 3106 (1982).
159. B. V. Derjaguin and D. V. Fedoseev, Jzd. Nauka (Moscow) (1977).
160. B. V. Derjaguin, L. L. Bouilov, and B. V. Spitsyn, *Arch. Nouki Mater.* **7**, 111 (1986).
161. T. Anthony, personal communication, 1982.
162. Y. Hirose, S. Amanoma, N. Okoda, and K. Komaki, Abstract, *First International Conference on New Diamond Science and Technology*, Tokyo, Japan, 1988, p. 38.
163. T. P. Ong and R. P. H. Chang, *Appl. Phys. Lett.* **55**(20), 2063–2065 (1989).
164. J. Suzuki, H. Kawarada, K. Mar, J. Wei, Y. Yokota, and A. Hiraki, *Jpn. J. App. Phys.* **28**(2), L281–L283 (1989).
165. R. Komanduri and S. Nandyal, *Int. J. Mech. Tools Manuf.* **33**(2), 285–296 (1993).
166. E. E. Sprow, *Manuf. Eng.*, 41–46 (Feb. 1995).
167. P. K. Bechmann, D. Leers, and H. Lydtin, *Diamond Relat. Mater.* **1**, 1–12 (1991).
168. R. Haubner and B. Lux, *J. de Phys.* **C5**(5), C5169–C5176 (May 1989).
169. U.S. Pat. 5,236,740 (Aug. 17, 1993), M. G. Peters and R. H. Cummings (to National Center for Manufacturing Sciences (NCMS)).
170. K. S. Reckling, *Tool Prod.*, **74** (Dec. 1981).
171. N. Tabuchi and co-workers, *Sumitomo Elec. Tech. Rev.* **18**, 57 (Dec. 1978).

172. P. M. Klutznick, *1981 U.S. Industrial Outlook for 200 Industries with Projections for 1985*, U.S. Department of Commerce, Washington, D.C., 1981.
173. G. Byrne and E. Scholta, *Ann. CIRP* **42**(1), 471–474 (1993).
174. *Machinery Prod. Eng.*, 14–20 (June 3, 1994).
175. R. B. Aronson, *Manuf. Eng.* **114**(1), 33–36 (Jan. 1995).
176. *Material safety data sheets*, Carboloy Company, Detroit, Mich.
177. M. E. Lichtenstein, F. Bartl, and R. T. Pierce, *Am. Ind. Hyg. Assoc. J.*, **879** (Dec. 1975).
178. *Turning Handbook of High Efficiency Cutting*, GT9-262, Carboloy Systems Business Department, General Electric Company, Detroit, Mich., 1980.
179. J. H. Crawford and M. E. Merchant, *Trans. ASME* **75**, 561–566 (May 1953).

RANGA KOMANDURI
Oklahoma State University

Related Articles

Steel; Carbides, cemented; Diamond; Boron compounds