

Chapter 4

Cutting Tool Materials and Tools

Tool change times, and with them both manufacturing times and tool, machine and labour costs, are affected by wear. Wear is affected in turn by the properties of the cutting tool materials. Development in the cutting tool material sector is therefore far from finished, but is constantly aiming both to improve cutting tool materials that are already established as well as to discover new materials for use in the manufacture of cutting tools.

Cutting tool materials should have the following properties in order to do justice to the stresses placed on them:

- hardness and pressure resistance,
- bending strength and toughness,
- edge strength,
- inner bonding strength,
- high temperature strength,
- oxidation resistance,
- small propensity to diffusion and adhesion,
- abrasion resistance,
- reproducible wear behaviour.

Should we consider all these characteristics simultaneously, we are then faced with the requirement for the “ideal” cutting tool material (Fig. 4.1). However, there is no one cutting tool material that unifies all the required properties. One reason for this, for example, is the physical opposition of hardness and toughness. Developments in the cutting tool material sector are concentrating on optimizing and modifying chemical composition, manufacturing methods, geometry and coating in order to broaden the areas of application of cutting tool materials and tools in accordance with the requirements of modern production.

4.1 Overview of Cutting Tool Materials

The cutting tool materials used in cutting can be summarized as (Fig. 4.2):

- tool steels,
- cemented carbides,

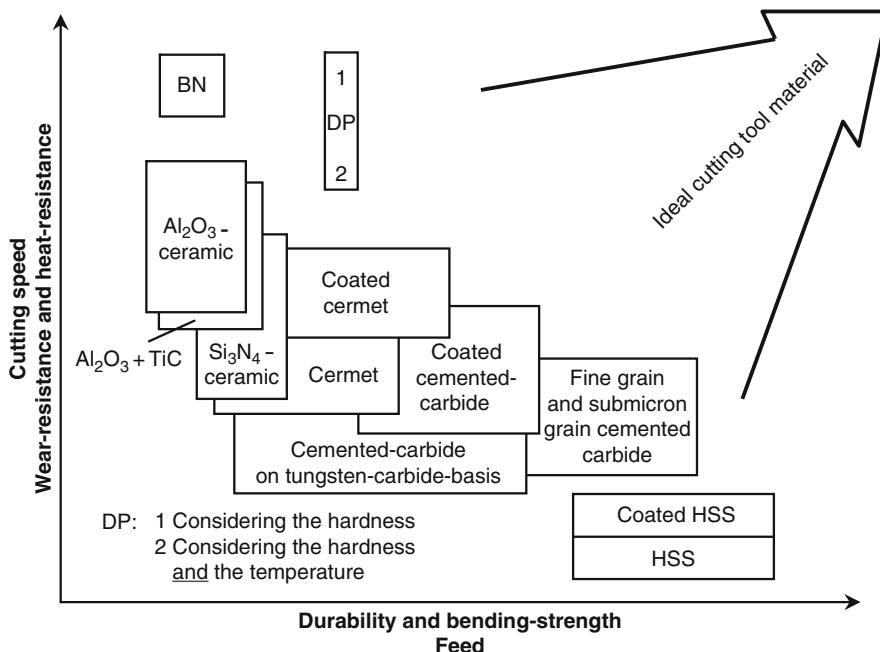


Fig. 4.1 Schematic classification of several cutting tool materials

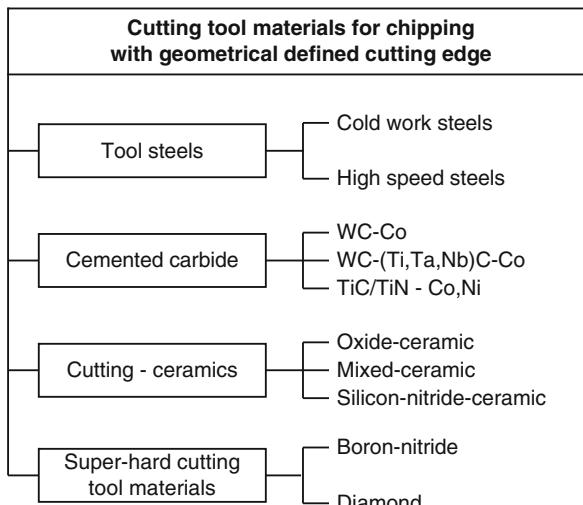


Fig. 4.2 Classification of cutting tool materials for machining

- cutting ceramics and
- super-hard cutting tool materials made of boron nitride and diamond.

The hardness and wear resistance of the cutting tool materials increases in this order. On the other hand, their bending strength decreases in this order.

Tool steels include unalloyed and alloyed cold work tool steels as well as high speed steel. The term cemented carbide refers to conventional cemented carbides based on tungsten carbide as well as “cermets” based on titanium boron nitride. Ceramics is the generic term for cutting tool materials made of oxide, mixed and non-oxide ceramic.

4.1.1 Classification of Hard Cutting Tool Materials

According to DIN ISO 513, the notation and application of hard cutting tool materials of cemented carbide, ceramic, diamond and boron nitride are classified in accordance with Fig. 4.3. According to it, uncoated cemented carbides based on tungsten carbide with a WC grain size of $\geq 1 \mu\text{m}$ take the abbreviation HW and with grain sizes of $< 1 \mu\text{m}$ the notation HF. “Cermets” – cemented carbides based on titanium boron nitride – are notated with HT, coated cemented carbides and coated cermets with HC. Corresponding abbreviations from Fig. 4.3 are valid for other hard cutting tool materials based on ceramic, diamond or boron nitride.

The goal of DIN ISO 513, besides that of identifying cutting tool materials, is above all their assignment to materials, for which they are most suitable for cutting. Expanding earlier norms, DIN ISO 513 provides six main application groups and thus six classes of workpiece-materials, which are classified with the code letters P, M, K, N, S and H as well as by colour (Fig. 4.4). DIN ISO 513 thus maintains the established code letters P, M and K but now only designate with them the main application groups that comprise the material groups steel (P), stainless steel (M), and cast iron (K). The major machining group K included according to the old standard not only cast iron materials but also a number of other materials. In DIN ISO 513, these now new major application groups were provided with the labels N for nonferrous metals, S for special alloys and H for hard materials.

Every main application group is subdivided into particular application groups (Fig. 4.4). These are notated with a code letter for the main application group they belong to and with an index. The index refers to the toughness and wear resistance of the cutting tool material. The higher it is within each application group, the lower is the material's wear resistance and the higher its toughness. The indices are merely reference numbers indicating a certain sequence. They provide no information about the extent of the wear resistance or the toughness of a cutting tool material. Manufacturers of cutting tool materials should assign their cutting tool materials to the appropriate application group, depending on wear resistance and toughness. Examples are HW-P10, HC-K20, and CA-K10. An application group thus contains comparable cutting tool materials of different manufacturers, although they may differ in wear behaviour and performance. It is also possible for a cutting tool material from one manufacturer to be categorized under multiple application groups if suitable to them.

Cemented carbide		Ceramic	
Code-letter	Group of materials	Code-letter	Group of materials
HW	Uncoated cemented carbide, predominantly made from tungsten-carbide (WC) with a grain size = 1 µm	CA	Oxide-ceramic, predominantly made from aluminium-oxide (Al_2O_3)
HF	Uncoated cemented carbide, predominantly made from tungsten-carbide (WC) with a grain size < 1 µm	CM	Mixed-ceramic, based on aluminium-oxide (Al_2O_3), but also with other components than oxides
HT ¹⁾	Uncoated cemented carbide, predominant titan-carbide (TiC) or titan-nitride (TiN) or both	CN	Silicon-nitride-ceramic, predominantly made from silicon-nitride (Si_3N_4)
HC	Cemented carbide like above, but coated	CR	Cutting-ceramic, predominantly made of Aluminium-oxide (Al_2O_3), reinforced
¹⁾ These cemented carbides are also called "Cermet"		CC	Cutting-ceramic like above, but coated

Diamond		Boron-nitride	
Code-letter	Group of materials	Code-letter	Group of materials
DP	Polycrystalline diamond	BL	Cubiccrystalline boron-nitride with a low content of boron-nitride
DM	Monocrystalline diamond	BH	Cubiccrystalline boron-nitride with a high content of boron-nitride

Diamond		Boron-nitride	
Code-letter	Diamond category	Code-letter	Boron-nitride category
DP	Polycrystalline diamond	BL	Cubiccrystalline boron-nitride with a low content of boron-nitride
DM	Monocrystalline diamond	BH	Cubiccrystalline boron-nitride with a high content of boron-nitride

Fig. 4.3 Notation of hard cutting tool materials for machining, acc. to DIN ISO 513

Main application group			Application group				
Code-letter	Code-color	Workpiece-material	Hard cutting tool materials				
P	blue	Steel: All kinds of steel and cast-steel, except stainless steel with austenitic structure	P01 P10 P20 P30 P40 P50	P05 P15 P25 P35 P45		↑	↓
M	yellow	Stainless steel: Stainless austenitic and austenitic-ferritic steel and cast-steel	M01 M10 M20 M30 M40	M05 M15 M25 M35		↑	↓
K	red	Cast-iron: Cast-iron with flake-graphite, cast-iron with ductile graphite, annealed cast-iron	K01 K10 K20 K30 K40	K05 K15 K25 K35		↑	↓
N	green	Nonferrous metals: Aluminium and other nonferrous metals, non-metal materials	N01 N10 N20 N30	N05 N15 N25		↑	↓
S	brown	Specializations and titanium: Highly heat resisting special-alloys based on iron, nickel and cobalt, titanium and titanium based alloys	S01 S10 S20 S30	S05 S15 S25		↑	↓
H	grey	Hard materials: Hardened steel, hardened cast-iron-materials, cast-iron for chill-casting	H01 H10 H20 H30	H05 H15 H25		↑	↓
<p>↑ Increasing cutting speed, increasing wear-resistance of the cutting tool materials</p> <p>↓ Increasing feed, increasing durability of the cutting tool materials</p>							

Fig. 4.4 Application and classification of hard cutting tool materials, acc. to DIN ISO 513

4.2 Tool Steels

DIN EN ISO 4957 defines tool steels as high-grade steels suited to machining and processing materials as well as handling and measuring workpieces. We distinguish between cold work steels, hot work steels and high speed steels. Cold and hot work steels are suited to purposes in which the surface temperature during engagement

is generally under 200°C. High speed steels on the other hand can be applied at temperatures of up to 600°C. Tools for both cutting and forming are made out of cold work steels and high speed steels. Hot work steels are used primarily for forming tools such as forging dies, pressure casting dies or tools for forging machines and extrusion presses.

4.2.1 Cold Work Steels

Cold work steels obtain their wear and toughness properties from heat treatment, consisting of heating to austenitizing temperature, quenching in an oil or water bath (a high cooling speed is necessary, martensite hardness) and tempering (with the goal of a partial loss of hardness in order to increase toughness).

Cold work steels can be classified as unalloyed and alloyed steels (Fig. 4.5) [DIN EN ISO 4957, Wegs95]. Unalloyed tool steels (carbon steels) contain up to 1.25% C

Unalloyed cold work steels							
Material grade Material number	Composition						Examples
	C	Si	Mn	Cr	Mo	V	W
C45W 1.1730	0.40 — 0.50	0.15 — 0.40	0.60 — 0.80				Hammers, hatchets, axes, scissors, screwdrivers, bits
C85W 1.1830	0.80 — 0.90	0.25 — 0.40	0.50 — 0.70				Wood saws, hand saws, steel blades for segment-buzz saws
C125W 1.1563	1.20 — 1.35	0.10 — 0.30	0.10 — 0.35				Files, ductors, chisels, paper knives
Alloyed cold work steels							
45CrMoV7 1.2328	0.42 — 0.47	0.20 — 0.30	0.85 — 1.0	1.7 — 1.9	0.25 — 0.30	— — 0.05	Chisel of all kinds
115CrV3 1.2210	1.10 — 1.25	0.15 — 0.30	0.20 — 0.40	0.5 — 0.8	— — —	0.07 — 0.12	Files, ductors, chisels, Paper knives
X210CrW12 1.2436	2.0 — 2.25	0.10 — 0.40	0.15 — 0.45	11.0 — 12.0	— — —	0.6 — 0.8	Cutting-tools, shear-blades for cutting of steel, broaches, wood-working-tools

Fig. 4.5 Examples for alloyed and unalloyed cold work steels

and small amounts of Si and Mn. Alloyed tool steels have about 1.25% C as well as up to 1.5% Cr, 1.2% W, 0.5% Mo and 1.2%V.

The hardness and wear resistance of unalloyed tool steels depend on their martensitic structure. Wear resistance increases with hardness and carbon content; yet toughness decreases simultaneously and thus the material's sensitivity during heat treatment and tool use becomes greater. All unalloyed tool steels are shell hardeners, i.e. they do not harden all the way through along the entire cross-section but only on the surface of the workpiece.

The advantages of alloyed tool steels in contrast to unalloyed are the increase in wear resistance (addition of carbide-forming elements), retention of hardness and high temperature strength (chrome, tungsten, molybdenum, vanadium alloys) and in their higher hardness (carbon in solution). Moreover, the critical cooling speed is lower, allowing for improved hardenability. They can be used at cutting temperatures of up to 200°C. They are used above all in steel machining with low cutting parameters (reaming, thread die cutting) and to manufacture tools for repair work, since their cost is lower than high speed steel (HSS) tools because they have fewer alloy elements.

Due to their low hot hardness, which limits the cutting speeds with which they can be used, cold work steels are used only rarely for metalworking on machine tools. Their area of application extends mainly to hand tools such as files, gouges, reamers or on saw blades for woodworking (Fig. 4.5).

4.2.2 High Speed Steels

A new cutting tool material was introduced for the first time in the year 1900 at the world's fair in Paris, with which TAYLOR could realize significantly high cutting speeds. The productivity of machining processes could thus be increased considerably. This new group of cutting tool materials was called "high speed steels" (HSS).

High speed steels are high-alloyed steels containing tungsten, molybdenum, vanadium, cobalt and chrome as their main alloy elements. They have relatively high bending fracture strength and thus have favourable toughness properties.

In contrast to cold work steels, their matrix is characterized by improved retention of hardness, and they have higher hardness. Their hardness of about 60–67 HRC is preserved up to 600°C in temperature. Due to both this and their machinability, they continue to have a broad range of application, especially for tools with sharp cutting edges and small wedge angles such as broaching tools, twist drills, thread-cutting tools, reamers, milling cutters and turning tools for grooving and parting-off operations as well as for finishing.

While the hardness of high speed steels are affected by the hardness of the base material (martensite) and the number and distribution of the carbides, the alloy elements dissolved in the matrix W, Mo, V and Co, which, partially precipitated as stable special carbides, are responsible for retention of hardness. Hardness and wear resistance are increased by the martensite tempered in the matrix and the embedded

carbides (especially Mo-W double carbides, Cr and V carbides). Carbide formation and hardening is promoted by additional alloying of chrome.

4.2.2.1 Classification of High Speed Steels

High speed steels are denoted with the letters "HS" and the percents of alloy elements in the order W-Mo-V-Co (e.g. HS10-4-3-10). High speed steels are classified according to their W and Mo content into four alloy and performance groups (Fig. 4.6) [DIN EN 10027a].

Group I includes the steels that contain high amounts of tungsten (18% W). The type containing Co HS18-1-2-5 has good retention of hardness. Tools made of this alloy are used for roughing high strength steels, cast irons that are difficult to machine, nonferrous metals and non-metallic materials.

The material HS12-1-4-5 of the group with 12% W has excellent wear resistance due to its high V content. Its Co content also gives it high hot hardness and retention of hardness. This alloy is used in the production of turning tools and profile steels of all kinds, for finishing tools, high performance milling cutters and automatic tool machines. They are suitable for machining highly heat-treated Cr-Ni steels and non-ferrous metals.

	Composition	Symbol W - Mo - V - Co	For conditioning of steel under...			
			...average load < 850 N/mm ²	> 850 N/mm ²	Roughing	Finishing
I	18% W	HS18-0-1	+	-	-	-
		HS18-1-2-5	-	-	+	-
II	12% W	HS12-1-[4]-5	-	-	(+)	+
		HS10-4-[3]-10	-	-	(+)	+
III	6% W + 5% Mo	HS6-5-2	-	+	-	-
		HS6-5-[3]	-	-	(+)	+
		HS6-5-2-5	-	-	+	-
IV	2% W + 9% Mo	HS2-9-1	+	-	-	-
		HS2-9-2	-	+	-	-
		HS2-10-1-8	-	-	+	-

Fig. 4.6 Alloy and performance groups of high speed steels

The last two groups are represented primarily by steels containing tungsten and molybdenum. Molybdenum replaces tungsten from metallurgical points of view and is more effective at equal masses because it has an approximately twice as large volume percentage due to its lower density. Steels containing molybdenum are especially tough. Types containing cobalt are used for machining with simple tools when robust stresses are expected (drills, turning, milling, planing and broaching tools, hobs). They are characterized by high toughness, temperature resistance and retention of hardness. They are universally applicable for roughing and finishing tasks, especially for highly stressed milling tools.

Steels of both of these groups that have low amounts of or no cobalt are used to manufacture tools of all kinds. The most important group are high speed steels with 6% tungsten and 5% molybdenum. Usually, the universal or standard type HS6-5-2 is used [Habe88]. As a result of its balanced alloy structure, its high toughness and good wear resistance, this alloy is very diverse. It is suitable for manufacturing machining tools for roughing or finishing such as twist drills, milling cutters of all types, tappers, reamers, broaches etc.

The basic effect of alloy elements in high speed steels is summarized once more in abbreviated form in the following:

- Tungsten: forms carbides; increases hot hardness, retention of hardness and wear resistance.
- Molybdenum: forms carbides; improves hardening and toughness, increases hot hardness, retention of hardness and wear resistance, molybdenum can replace tungsten (has half its density!).
- Vanadium: exists as the primary carbide VC and increases wear resistance (finishing).
- Cobalt: shifts the limit of overheating sensitivity to higher temperatures, making it possible to achieve higher hardness temperatures. Most carbides are dissolved and the hot hardness increases.
- Chrome: improves hardenability, participates in carbide formation.
- Carbon: the source of hardness in the base material. Increases wear resistance because it forms carbides.

4.2.2.2 Areas of Application

Figure 4.7 provides a list of the main applications of high speed steels [DIN EN ISO 4957, Wegs95]. As the amount of alloy elements increases, so does the effectiveness of these cutting tool materials with respect to improved wear resistance and lifetime. At the same time however, machinability becomes more difficult, which has negative consequences particularly when manufacturing complicated moulding tools. Generally, a higher amount of alloy elements means higher tool costs. The cost effectiveness of a manufacturing process is thus also determined by the choice of high speed steel.

This use of high-alloyed high speed steels lends itself especially to the solution of machining problems in which an increase in high temperature strength or

Steel grade name according DIN EN ISO 4957	Material no.	Main use
HS6-5-2	1.3343	Standard high speed steel for all cutting tools for rough turning or finish turning, taps and twist drills, milling cutters of all kinds, pull broaches, reamers, threading dies, counterboring drills, planing-tools, buzz saws
HS6-5-3	1.3344	Highly loaded taps and reamers, high speed milling cutters, broaching-tools, twist drills, cutting and shaving wheels
HS6-5-2-5	1.3243	High speed milling cutters, planing- and turning-tools of all kinds, highly loaded twist drills and taps, broaches, woodworking- and cold working-tools, roughing-tools with high durability
HS10-4-3-10	1.3207	All-purpose usage for roughing- and finishing-works, turning- and highly loaded milling cutters, automated works, woodworking-tools
HS2-9-2	1.3348	Milling cutter, reamers, broaching-tools
HS2-9-1-8	1.3247	End-milling cutter, turning-tools for automated works, twist drills, taps

Fig. 4.7 Main applications of the most important high speed steels, acc. to DIN EN ISO 4957

toughness would have a great effect. High speed steels alloyed with cobalt (e.g. HS6-5-2-5, HS18-1-2-5) are ideal for machining tasks that place increased requirements on the high temperature strength of the tools. Steels that contain vanadium in addition to cobalt, such as the qualities HS12-1-4-5 and HS10-4-3-10, are characterized by improved wear properties at increased temperatures stresses and are suited to machining tasks that place the highest requirements on the wear resistance of the tools.

4.2.2.3 High Speed Steel Fabrication

High speed steels can be fabricated using cast-metallurgical [Habe88] or powder-metallurgical methods [Duda86, Beis82]. A comparison of both fabrication methods is provided in Fig. 4.8. Despite attempts to develop alternative fabrication techniques, the cast-metallurgical fabrication of high speed steels remains the most economically important.

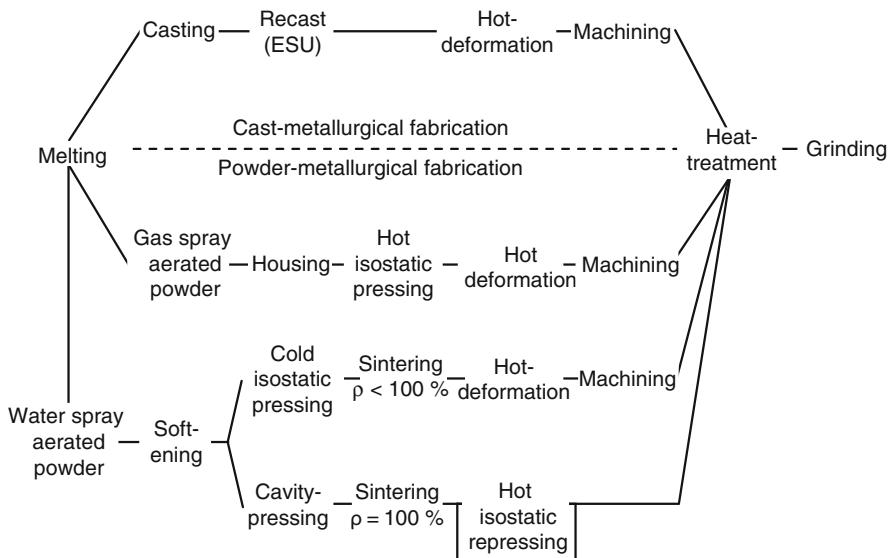


Fig. 4.8 Production-sequence of the powder- and cast-metallurgical fabrication of high speed steels

Cast-Metallurgical Fabrication

After melting and casting (1550°C) the steel in moulds, it is then subjected to block tempering (900°C) (homogenization, Fig. 4.9). This is followed by a forging process (to break up the ledeburite and carbides) with a possible intermediate heating of the workpiece, as well as by rolling (1200°C). Softening improves machinability. The softened condition is in most cases the post practical both for cutting and cold forming and offers the most favourable output structure for hardening. In order to avoid cracks or warpage during hardening, it is advisable to anneal the tools with a minimal amount of stress in order to reduce machining stresses before the last mechanical processing, especially in the case of tools that are irregular or difficult to mould.

High speed steels require a highly complex hardening sequence due to the high amounts of different alloy elements. Conventionally produced high speed steels thus tend to segregate (demix) during the hardening phase, which is especially pronounced in the case of large cast blocks and high-alloyed steels. Such phenomena are usually causes of lifetime deviations in later tool use.

Independently of previously applied measures such as improved melting process or seeding as well as improved design of the block format, electro slag remelting (ESR) can further improve the degree of purity and toughness of high speed steels. In this process, only a small part of the block is fluid during remelting, not the whole block. In this way, macroscopic demixings such as carbide agglomerations can be reduced and slag inclusions removed. The disadvantage is the higher fabrication cost.

Code = Grade	HS18-1-2-5	HS10-4-3-10	HS12-1-4-5	HS12-1-4	HS6-5-2
Hardening temp. [°C]	1280	1240	1240	1240	1230
Annealing, each 0.5–1 h	3 × 560 °C	2 × 570 °C 1 × 550 °C	2 × 570 °C	2 × 560 °C	2 × 540 °C

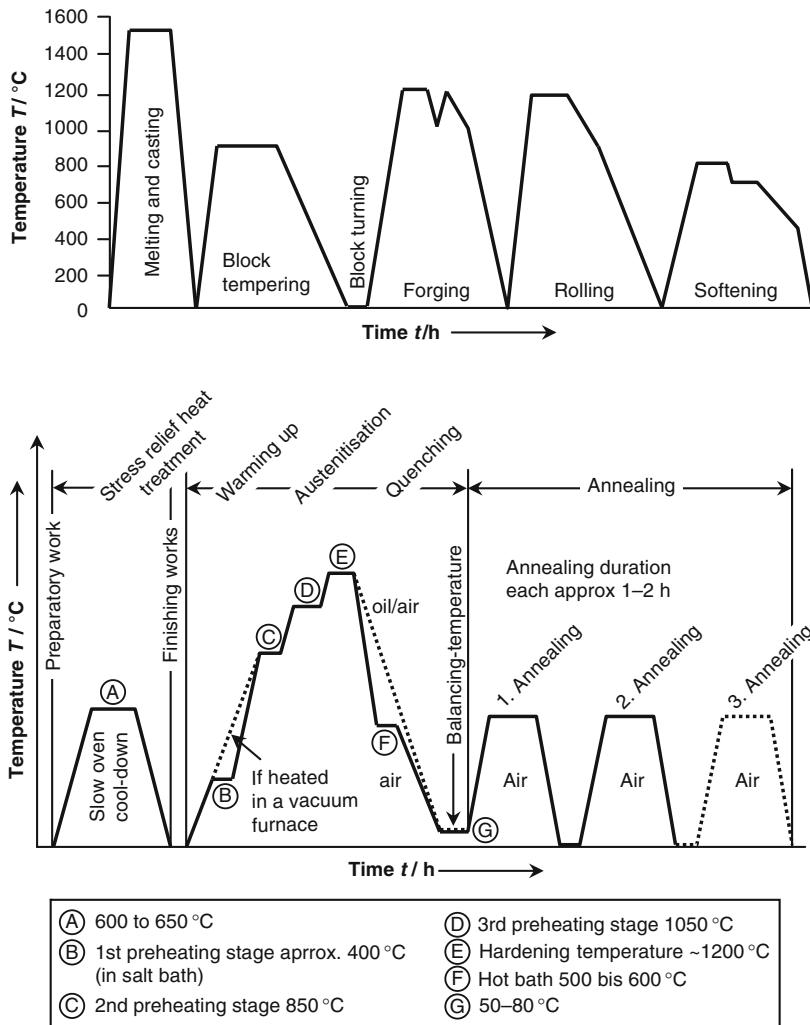


Fig. 4.9 Cast-metallurgical fabrication and heat treatment of high speed steels

The advantage of the ESR process is that the end structure is more even, has a small amount of segregations and a high degree of purity. The practical consequences of this are reduced warpage in both transverse and longitudinal directions during hardening and annealing due to the more homogeneous microstructure – an especially important aspect in the case of tools that are long or are premachined with a small stock allowance – as well as improved toughness resulting from the higher degree of purity and the generally more homogeneous structure.

Powder-Metallurgical Fabrication

The starting materials for powder-metallurgical fabrication of high speed steels is powder obtained by means of gas or water spray aeration of the molten mass (Fig. 4.8). Gas spray aerated powder is housed and undergoes hot isostatic pressing. After soft deformation, the semi-finished products are conventionally machined into tools. Water spray aerated powder undergoes a cold isostatic pressing after softening, and then the semi-finished products are sintered in a vacuum furnace. Sintering is not performed until the full theoretical density is reached. Final pressing takes place later in a forging process downstream. On the other hand, mouldings such as indexable inserts are hardened in matrix presses. During sintering, the aim is to reach the full theoretical density. The sintering process can sometimes be followed by a hot isostatic repressing.

Powder-metallurgically obtained high speed steels have become increasingly important in recent years. As opposed to cast-metallurgically fabricated high speed steels, they generally have a somewhat higher alloy content. PM steels are characterized by a homogeneous structure (no carbide segregation) with an even distribution of fine carbides. Due to their structural composition, PM steels have better grindability and higher toughness. With respect to their efficiency as cutting tools, powder-metallurgically fabricated high speed steels vary to some degree. Numerous cutting experiments have shown that they are at least equal to conventional high speed steels with the same nominal composition. PM steels become advantageous under high mechanical stress resulting from greater feeds, especially when cutting tool materials that are difficult to machine, such as nickel-based and titanium alloys.

Until now, powder-metallurgically fabricated high speed steels have no material indexes of their own. They are designated according to the company that manufactures them. To simplify their classification, the American nomenclature is often used, or the standard designation of the corresponding cast-metallurgically fabricated high speed steel is applied.

4.2.2.4 Heat Treatment of High Speed Steels

HSS tools are subjected to heat treatment in order to give them their required hardness [Habe88]. First they are hardened (heating and maintaining an austenitisation temperature followed by high-speed cooling) and then annealed several times. Figure 4.9 shows a time-temperature chart for the heat treatment of high speed steel HS6-5-2.

Due to their poor heat conductivity, high speed steels must be heated very slowly in order to avoid heating stresses and cracks. Warming up thus generally takes place in two or three preheating stages at 400, 850 and 1050°C. After reaching the level of austenitisation, a temperature of 1150–1250°C is held for hardening. The goal of austenitisation is to dissolve as many carbides as possible, whereby the dissolution of the secondary carbides increase with a rising hardening temperature and extended holding times. On the other hand, coarseness must also be avoided. Depending on the critical quenching speed and on the shape and size of the work-pieces, quenching can take place in different media, including water, oil, hot baths (salt or metal molten baths) or gas. Gaseous media include air that is still or in motion, nitrogen or other gasses. Quenching intensity, which in the case of gasses is much lower than with liquid media, can be increased by raising the flow speed and pressure.

After quenching, the microstructure of high speed steels consists of martensite, residual austenite and carbides. The hardness of this structure is unsuitable for cutting tools; due to the high amount of residual austenite and the instability of the structure, it has insufficient wear resistance, is not dimensionally stable during use and exhibits an increased fracture tendency. The tools must therefore undergo annealing after hardening.

High speed steels are annealed at temperatures of 540–580°C, the temperature range of secondary hardness. This is defined as the increase in hardness at higher annealing temperatures after an initial decline in hardness from annealing at low temperatures. Under normal hardness and annealing conditions, this second increase in hardness leads to hardness values that are clearly higher than those existing after quenching. Figure 4.10 is a schematic depiction of the processes that occur during the annealing of high speed steels. The cutting tool material hardness resulting

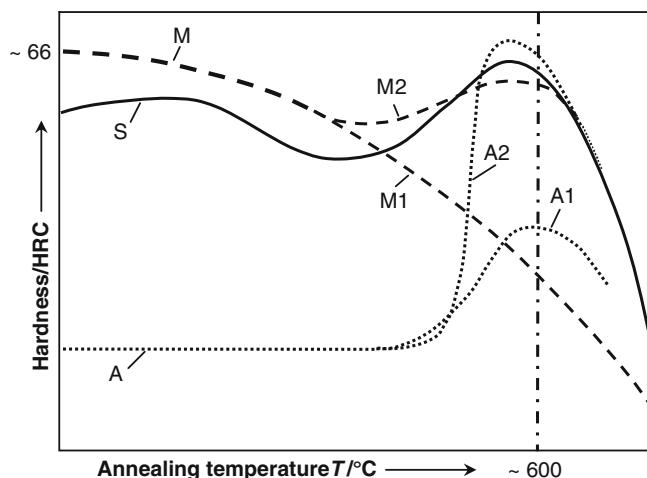


Fig. 4.10 Schematic illustration of cutting tool material hardness resulting from the superposition of various effects during the annealing of high speed steel

from the superimposition of various effects is plotted as the cumulative curve S as a function of the annealing temperature.

As the annealing temperature increases, the hardness of martensite (M) decreases continuously (M1). The cause of this is precipitation of the carbon trapped in solution and the associated reduction of crystal lattice tension. The precipitation of the carbides acts against this process, which leads to the decline in hardness observable during annealing at up to about 350°C. Extremely finely distributed carbides precipitate from the martensite (M) above 350°C and from the residual austenite (A) above around 450°C, causing another increase in hardness (M2/A1) due to precipitation hardening. As a result of carbide formation, the matrix becomes low in carbon and alloy elements, so that part of the residual austenite is converted to secondary martensite (A2) during cooling from annealing temperature. All three processes – carbide precipitation from the martensite, carbide precipitation from the residual austenite and the conversion of residual austenite into secondary martensite – lead to the increase in hardness known as the secondary hardness effect. The maximum of the hardness profile is shifted to higher temperatures (560°C), and, if austenitisation occurs correctly, hardness can be higher after annealing than in the hardened state [Habe88].

Until now, heat treatment usually took place in salt baths. Their advantages are optimal heat transfer and the possibility of executing partial heat and surface treatments. The important disadvantages are the high costs involved in detoxifying the salts and the necessity of cleansing the workpieces of salt residue after treatment.

The last few years have seen increasing amounts of heat treatments taking place in vacuum furnaces. The advantages of this method are above all its environmental friendliness as well as the clean and shiny workpiece surfaces it yields. Quenching the workpieces from hardening temperature is achieved by flooding them with nitrogen at a pressure of 10 bar. The gas is circulated such that it flows at high speed through the batch and then through a heat exchanger for recooling. Components are quenched in this manner to 50°C.

4.2.2.5 Surface Treatment

The wear properties of HSS tools can be further improved by treating the surface by nitrating (enriching the surface layer with nitrogen by thermochemical treatment at temperatures of 500–580°C), steam annealing (formation of a thin iron oxide layer on the tool surface in a steam atmosphere at about 500°C), chrome-plating (deposit of a hard chrome layer with a thickness of 5–50 µm at temperatures of 50–70°C) or coating (PVD coating with thin, highly wear-resistant hard material coats such as titanium nitride (TiN) or titanium aluminium nitride (Ti,Al)N).

4.2.2.6 HSS Indexable Inserts

Indexable inserts made of high speed steel can be manufactured by precision casting, by machining pre-products or directly using powder-metallurgical methods.

Profile milling cutter for rotors

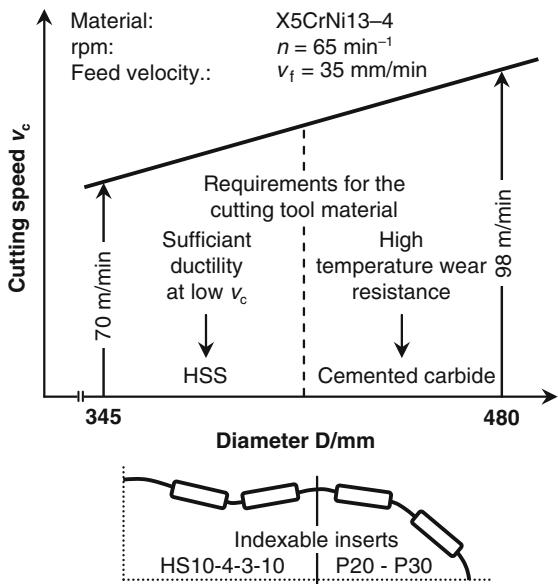
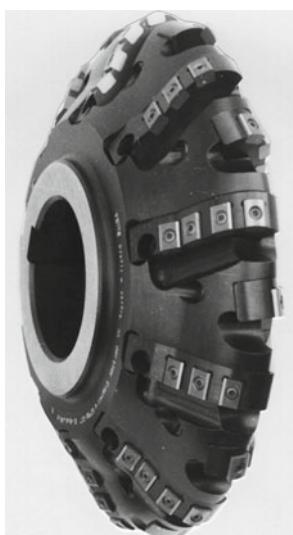


Fig. 4.11 Fit-for-purpose design of indexable inserts composed of high speed steel and cemented carbide (Source: Fette)

While pre-products still require further machining, powder-metallurgical fabrication of HSS indexable inserts for example exploit the cost-intensive alloy elements much more fully. Besides lower manufacturing costs and more efficient utilization of material, the most important advantage of sintered HSS indexable inserts is their increased flexibility in tool use due to the use of indexable insert technology. In the case of manufacturing processes such as milling or drilling, this technology allows us to fit tools with various fit-for purpose cutting tool materials (Fig. 4.11). Indexable inserts made of precision casting have so far gained no importance for steel machining [Bong91, Wähl86].

4.3 Cemented Carbides

Cemented carbides are composite materials. They are composed of the carbides of transition metals (fourth to sixth subgroup of the periodic system), which are embedded in a soft metallic binder phase made of cobalt and/or nickel. The carbides are on the boundary between metals and ceramics. They still have metallic properties (e.g. electrical conductivity), but they are assigned to the non-oxide ceramics as “metallic hard materials” [Horn06, Salm83]. Additionally, cermets contain titanium nitride.

Hard materials are the bearers of hardness and wear resistance. The binder phase is responsible for binding the brittle carbides and nitrides to a relatively solid body.

The advantages of cemented carbides include good structural uniformity due to their powder-metallurgical fabrication, high hardness, pressure resistance and high-temperature wear resistance. At 1000°C, cemented carbides have the same hardness as high speed steel at room temperature. It is also possible to manufacture types of cemented carbides with various properties by means of intentionally changing the amounts of hard material and binder [Sche88, Kola92].

4.3.1 Historical Development

In 1927, cemented carbides were first introduced as new high performance cutting tool materials at the Leipzig Trade Fair under the name WIDIA (Fig. 4.12). This was a revolutionary development at the time, opening up completely new dimensions in cutting technology. In contrast to the high speed steel tools used hitherto, cemented carbides allowed the use of twice or three times the cutting speed with the same cross-section of undeformed chip. The capacities of the machine tools of the time were far beneath the potential of the new cutting tools. Materials such as chilled cast iron, which had been very difficult to cut with HSS tools, could be machined easily with the new cutting tool material. Initially, WC-Co cemented carbides were used exclusively to machine cast iron materials. Due to the high amount of crater wear, these cemented carbides were not suited to machining long-chipping steel materials. This changed with the introduction of titanium carbide as an alloy component. With cemented carbides containing TiC, similar sensational cutting speed increases were made possible in steel machining as previously was the case when machining cast iron with WC-Co cemented carbides. This period also saw the development of the first cermet. Cemented carbides made for patent reasons without tungsten based on TiC and Mo₂C with Nickel as the binder proved in many cases to be

1923	Patent for the fabrication of WC-Co-HM, K. Schröter
1927	Leipzig Trade Fair: 1st WC-Co-HM, producer Krupp, labeling WIDIA
1928	Carboloy (General Electric)
1931	WC-TiC-Co cemented carbide
1934	TiC-Mo ₂ C-Ni (1st generation of Cermets) producer Plansee, labeling Titanit S
1937	Titanit (Plansee), Böhlerit (Böhler)
1942	Coromant (Sandvik)
68/69	Implementation of coated cemented carbide
1970	Fine grain cemented carbide
1973	Spinodal - cemented carbide (1st Cermet with TiN)
since 73/74	Accelerated developing of Cermets
1993	Submicron grain and nano crystalline cemented carbide

Fig. 4.12 Milestones in the development of cemented carbides

excessively fragile for the machinery of the time [Sche88, Kola92, Kola93, DRP23, ÖP31, Häus90, Spri95].

Further development of cemented carbides in the following years led to continuous improvement of their composition, production and cutting performance. The influence of carbide grain size on the properties of cemented carbides was recognized early on. The relation, whereby cemented carbides can be increased in hardness only with a reduction in toughness, could be overcome with the development of fine grain cemented carbide. By reducing WC crystallite size to under $1\text{ }\mu\text{m}$, both hardness and bending strength could be increased with the same amount of binder [Sche88, Kola93, Spri95].

The introduction of coated cemented carbides at the beginning of the 1970s was another great innovation. The combination of tough cemented carbide substrates with highly wear-resistant hard material coats led to an enormous increase in possible cutting speeds and tool standing times. The CVD and PVD methods are the most important coating process variants today. Multi-layer coats on substrates with binder-rich rim zones or gradient structures are new developments [Sche88, Kola92, Gil95].

With the development in 1973 of “spinodal” cemented carbides, the first Cermet, which contained titanium nitride as a further hard material component, the basic form of today’s highly efficient cermet was created. Cermets are today among the high performance cutting tool materials that meet the demands of modern cutting technology excellently by allowing for the use of high cutting speeds with moderate feeds and realizing long standing times with a high level of reliability. This is due to their high chemical stability and high-temperature wear resistance, making these cutting tool materials especially interesting for cutting operations with high thermal stress on the cutting edge.

4.3.2 Cemented Carbide Production

Cemented carbides are fabricated using various powder-metallurgical means (Fig. 4.13). Due to the great variety of shapes that cemented carbide components can assume, very different forming methods can be used. The most general distinction is between direct and indirect fabrication, combined fabrication and special methods (injection moulding). The fabrication method used depends mainly on the geometry and quantity of the product to be manufactured [Sche88, Kola92].

About two thirds of all cemented carbide products are manufactured by direct fabrication, primarily indexable inserts. Complicated moulding in small quantities, such as pistons, screws, rolling rings or matrices are made indirectly, i.e. using additional machining steps such as cutting, drilling, turning or milling. The starting material is a cemented carbide in a pre-sintered or cold-isostatically pressed condition, the consistency of which is still chalky [Sche88, Kola92].

The individual components of the cemented carbide are weighed out as powder to a batch and homogenized in mixers. In the wet grinding following this, the grinding fluid (alcohols, acetone, hexan) should protect the powder from oxidation during

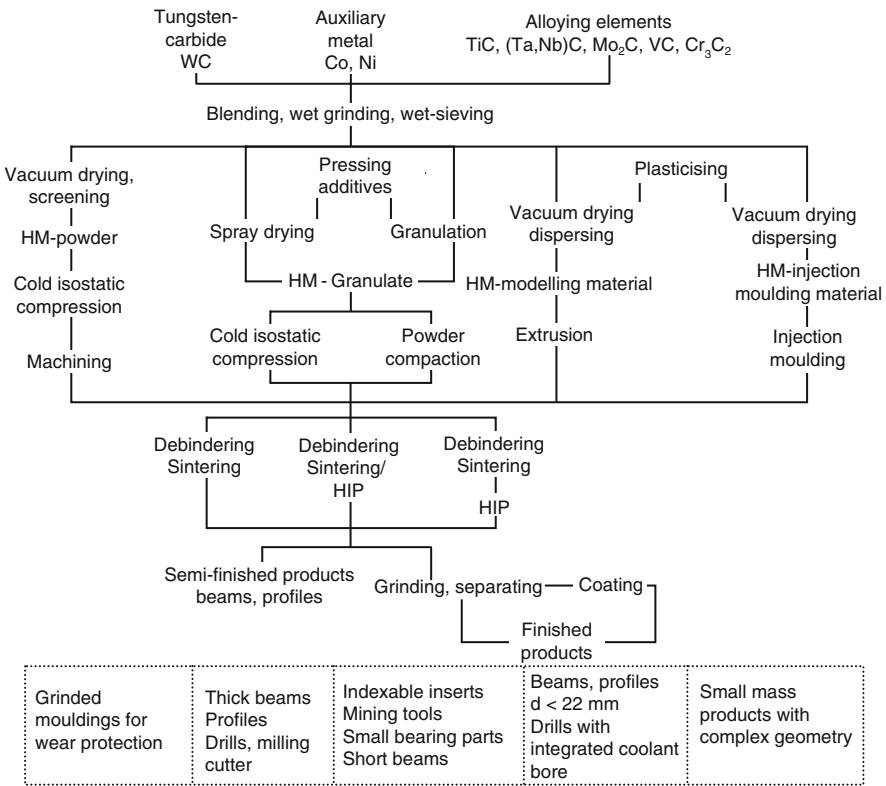


Fig. 4.13 Fabrication sequences for the production of workpieces made of cemented carbide (Source: Widia)

grinding and guarantee optimal dispersal of all the components in the suspension. After the grinding process is finished, the powder mix is prepared in accordance with the subsequent shaping method. To produce indexable inserts, the powder is transformed into granulate in order to assure good flow properties and a suitable granulate size for compression into matrices. The granulate is produced with the help of spray drying or granulation processes. Shaping the indexable inserts is achieved by compressing the granulate in matrix presses. The matrix press method allows for short cycle times and is thus especially suited to producing large quantities [Kola92, Daub95].

Within the fabrication sequence, sintering process is probably the most important operation, since it is here that the component obtains the mechanical and technological properties that are essential for its functional capability. In principle, sintering is defined as a thermally activated material transport in which the compressed, loosely bonded powder material (pressed part, green compact) is increasingly compressed due to diffusion-controlled place-shifting processes (surface diffusion, grain

boundary diffusion, volume diffusion). It is decisive for fabricating cemented carbides that not the entire alloy system is converted to a molten state, but rather (as it is called in professional jargon) sintered with liquid phase [Sche88, Kola92].

Molten baths first appear at about 1300°C. Tungsten carbide dissolves increasingly in the molten bath. At a sintering temperature of about 1400°C, the entire binder phase, composed of cobalt and tungsten carbide, is molten. The liquid phase wets the carbide, penetrates into all the pores and causes the hard material particles to glide together under the influence of surface tension into the smallest space. This particle rearrangement and re-precipitation process leads to tight packing with minimal surface energy. Depending on the green density and sintering parameters, linear shrinkage can reach 20%. As the mass cools from sintering temperatures, tungsten carbide precipitates again from the liquid binder phase. Up to a temperature of 700°C, a further thermally activated material transport of tungsten and carbon takes place in accordance with the solubility of WC in Co. Atoms of the hard material phase still dissolved after solidification in the binder phase stabilize the cobalt-rich binder phase in the cubic crystalline crystal lattice structure until room temperature; otherwise the cobalt converts into hexagonal form below 417°C [Sche88, Kola92].

4.3.3 Components of Cemented Carbides and Their Properties

- WC: Mono tungsten carbide is the most important hard material phase in technical sintered cemented carbides. WC is soluble in Co, resulting in the high inner bonding and edge strength of WC-Co cemented carbides. WC is also even more wear-resistant than TiC and TaC. The applicable cutting speed is limited at higher temperatures due to its tendency to dissolution and diffusion.
- TiC: Titanium carbide has a low tendency to diffusion. The consequence of this is that TiC cemented carbides have considerable high-temperature wear resistance but little bonding and edge strength. Cemented carbides high in TiC are therefore brittle and fragile. Their use is preferred when cutting steel materials with high cutting speeds. TiC can join with WC to make a composite carbide.
- TaC: In small amounts, tantalum carbide has a grain-refining effect, thus improving toughness and edge strength; the inner bonding strength does not decrease as sharply as is the case for TiC.
- NbC: NbC has a similar effect as TaC. Both carbides appear as a mixed crystal (Ta, Nb)C in cemented carbides.
- TiN: Titanium nitride is the property-determining component in all modern cermets. TiN in steel is even less soluble and is thus more resistant to diffusion than titanium carbide. Nitrogen causes an increase in wear resistance. In addition, grain growth is inhibited. Cermets containing nitrogen have as a rule a very fine-grained structure. In a solid state, TiC and TiN are completely mixable. The physical properties of cermets are based on those of titanium carbon nitride.
- Co: Cobalt is still unrivalled as a binder metal for cemented carbides based on tungsten carbide. This is due to the high level of solubility of WC in cobalt and

to the good wettability of tungsten carbide crystals by the molten WC-Co binder phase [Sche88].

- Ni: due to the improved wettability of hard materials, nickel is used as a binder for cermets. But since nickel is more easy to deform than cobalt, cobalt is also added with nickel as a binder in cermets today in order to improve their high temperature properties.

The properties of the cemented carbide substrate are of key importance with respect to the wear resistance and efficiency of uncoated and coated cemented carbides. Cemented carbides should have high hot hardness and pressure resistance for the sake of increasing the resistance of the cutting edge to plastic deformation. They should however also have high bending strength and therefore sufficiently high toughness and resistance to cracking, crumbling and fracture. In general, tough cemented carbides are low in hardness and pressure resistance. These opposing cutting tool material properties are influenced decisively by the microstructure of the base material. The tendency is that fracture toughness increases with the cobalt content and average grain size, while hardness and pressure resistance are decreased (Fig. 4.14). As the concentration of composite carbides goes up, fracture toughness is reduced. Tantalum carbide has a favourable effect on temperature change strength.

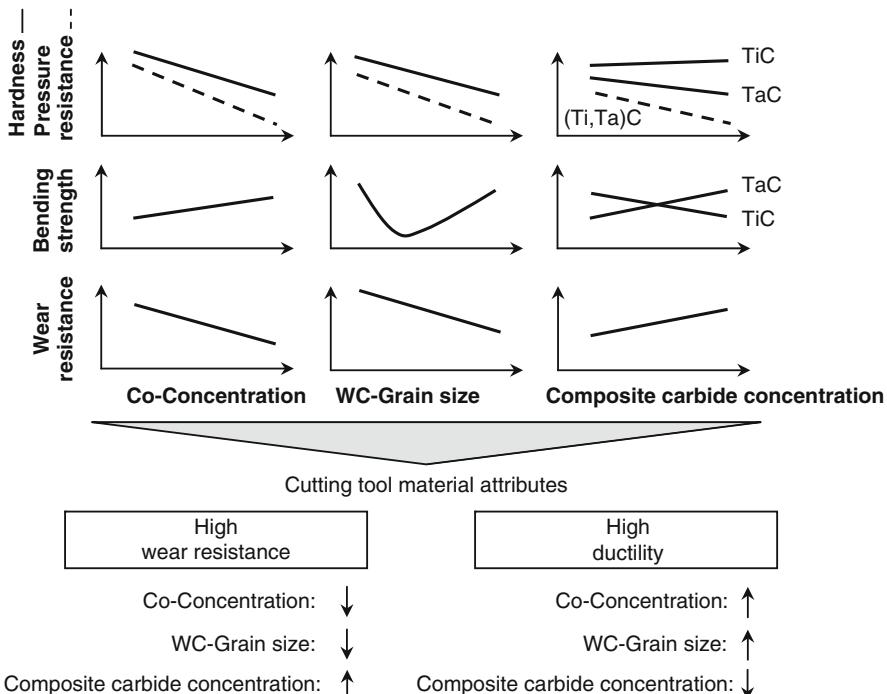


Fig. 4.14 Influencing factors for wear resistance of cemented carbides

This is exploited in many cases and, in the case of substrates for milling, the ratio TiC/TaC is modified in favour of tantalum carbide [Sche88, Schi89, Koni90]. The high material costs put a tight limit on the use TaC alloys however.

4.3.4 Microstructure

In conventional cemented carbides based on WC, the tungsten carbide is usually in the form of prisms with triangular bases. As opposed to cemented carbides containing larger amounts of cobalt, in which crystal growth is less hindered, this crystal form is less well developed in types that are fine-grained and low in cobalt. Cubic mixed crystals exist both in both cubical (with rounded edges) and nearly spherical shape (Fig. 4.15). The carbide skeleton is filled with the binder phase.

The structure of cermet contains only rounded composite carbon nitrides. The core/shell structure of the hard materials is characteristic of their microstructure. The causes of this are demixing phenomena in the hard materials (spinodal demixing), mainly however the dissolution and re-precipitation processes taking place during liquid phase sintering. By selective solution, the Ni-Co molten bath is enriched with carbide components. Composite carbides precipitate as a rim zone around the remaining hard material grains when cooling down from sintering temperature [Mosk66, Rudy73, Kola89, Kief71, Leop87, Kola89a, Kola93a].

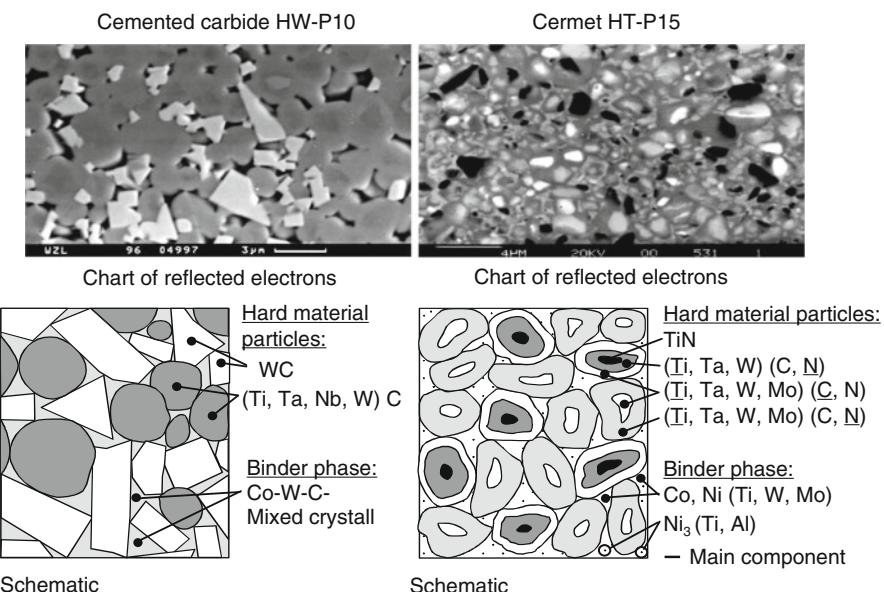


Fig. 4.15 Microstructure of conventional cemented carbides and cermets

The microstructure of the hard material particles is determined by the components contained in the original powder. The hard materials can exhibit widely varied textural structures depending on composition and grain size. If the initial powder consists of unalloyed binary hard material components, usually titanium nitride or titanium carbide are in the cores (dark cores). When prealloyed powders are used consisting of composite carbides or quaternary carbon nitrides, higher concentrations of molybdenum and/or tungsten are also found in the cores (light cores). The back scatter electron image in Fig. 4.15 shows this varying core/shell structure very clearly. The initial powder consists in this case of single carbides, composite carbides and carbon nitrides. The hard material particles with the dark cores have $\text{Ti}(\text{C},\text{N})$ in the centre, which is surrounded by a relatively titanium-rich composite carbon nitride $(\text{Ti},\text{Ta},\text{W})(\text{C},\text{N})$. The lighter $(\text{Ti},\text{Ta},\text{W},\text{Mo})(\text{C},\text{N})$ shell consists primarily of a titanium-tungsten-molybdenum composite carbide containing a relatively large amount of tungsten and/or molybdenum. The hard material particles in the light core contain this composite carbide inside, surrounded by a titanium and nitrogen-rich shell.

4.3.5 Classification of Cemented Carbides

Cemented carbide cutting tool materials can be subdivided into three groups. These are cemented carbides based on:

- WC-Co,
- WC-(Ti,Ta,Nb)C-Co and
- TiC/TiN-Co,Ni.

The term “cermet” is now used for cemented carbides based on TiC/TiN-Co,Ni.

4.3.5.1 WC-Co Cemented Carbides

The cemented carbides of this group consist almost exclusively of hexagonal tungsten monocarbide and the binder phase cobalt. They can contain up to 0.8 mass % VC and/or Cr_3C_2 and/or up to 2 mass % $(\text{Ta},\text{Nb})\text{C}$ as doping additives to control structural fineness and consistency (Fig. 4.16).

WC-Co cemented carbides are characterized by high abrasion resistance. Due to the strong diffusion tendency of tungsten carbide, they are not suitable for machining soft steel materials. They are mostly used for short-chipping materials, cast iron materials, non-ferrous metals and non-metals, high temperature materials as well as in stone and wood processing (Fig. 4.4).

According to previous terminology, WC-Co cemented carbides were subdivided in accordance with the average WC grain size in the sintered microstructure into fine ($0.8\text{--}1.3 \mu\text{m}$), submicron ($0.5\text{--}0.8 \mu\text{m}$) and ultrafine ($0.2\text{--}0.5 \mu\text{m}$). Deviating from this, the standard DIN ISO 513 distinguishes only between cemented carbides with grain sizes $\geq 1 \mu\text{m}$ (HW) and those with grain sizes of $< 1 \mu\text{m}$ (HF).

Cemented carbide Application group (DIN ISO 513)	HW-K05	HW-K10	HW-K25	HW-K40
Grade	WC-4Co	WC-6Co	WC-9Co	WC-12Co
Density (g/cm ³) (ISO 3369)	15.1	14.9	14.6	14.2
Hardness HV 30 (ISO 3878)	1730	1580	1420	1290
Pressure resistance (cylinder test) (N/mm ²) (ISO 4506)	5700	5400	5000	4500
Bending strength (N/mm ²) (ISO 3327)	1600	2000	2350	2450
Young's modulus (10 ³ ·N/mm ²) (ISO 3312)	650	630	590	580
Fracture toughness (N·m ^{1/2} /mm ²)	6.9	9.6	12.3	12.7
Poisson-constant	0.21	0.22	0.22	0.22
Heat conductivity (W·m ⁻¹ ·K ⁻¹)	80	80	70	65
Coefficient of thermal expansion (293–1073 K) (10 ⁻⁶ ·K ⁻¹)	5.0	5.5	5.6	5.9

Fig. 4.16 Composition and attributes of conventional WC-Co cemented carbides, acc. to Kolaska [Kola92]

Conventional Fine-Grain Cemented Carbides (HW)

Conventional uncoated WC-Co fine-grain cemented carbides (fine cemented carbides) with an average grain diameter of 0.8–1.3 µm (acc. to DIN ISO 513 grain size $\geq 1 \mu\text{m}$) are still widely used in cutting technology applications where high demands are placed on cutting edge sharpness and toughness, e.g. in steel milling, finishing, grooving and parting-off or in the manufacture of threads [Kola92].

Submicron and Ultrafine Cemented Carbides

Submicron and ultrafine cemented carbides have until now been the terms used in literature and practice for WC-Co cemented carbides with an average WC grain diameter of 0.5–0.8 and 0.2–0.5 µm in the sintered microstructure. The small grain diameter gives these cemented carbides a special combination of attributes: reducing the WC crystallite size below 1 µm with equal binder content leads both to increased hardness and bending strength (Fig. 4.17). This property makes submicron and ultrafine cemented carbides applicable to a wide variety of tasks [Kola92, Daub95, Köni90a, Köni93, Drey01, Gill01].

Valuable submicron and ultrafine cemented carbides are superior to conventional fine-grain cemented carbides in hardness, edge strength and toughness. Moreover, they have a small inclination to adhesion and to wear by diffusion. These attributes

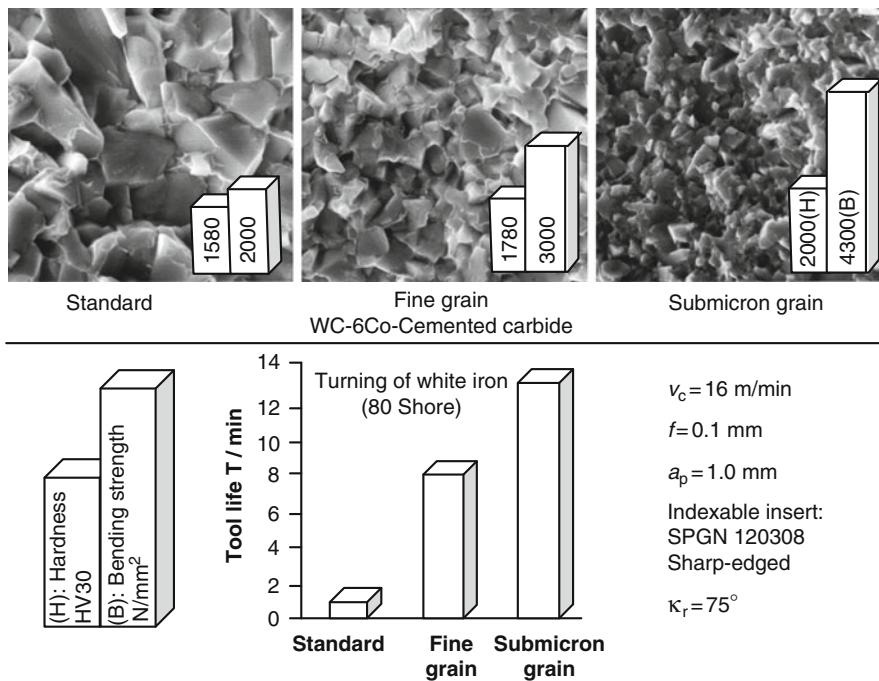


Fig. 4.17 Microstructure and attributes of fine and submicron grain cemented carbides in comparison to standard K-class cemented carbides (Source: Widia)

are necessary when hardened materials must be machined with finishing quality and the smallest machining allowances.

Submicron and ultrafine cemented carbides are used whenever high toughness, high wear resistance and the highest edge strength is required of the cutting edge, e.g. for broaching, milling and shaping of heat treated and hardened steels, in cast iron machining and for machining fibre-reinforced plastics and non-ferrous metals [Kola92, Daub95, Köni90a, Köni93, Drey01, Gill01].

4.3.5.2 WC-(Ti,Ta,Nb)C-Co Cemented Carbides

The cemented carbides of this group contain, besides tungsten carbide, mixed carbides (MC) made of titanium, tantalum, niobium and/or zirconium carbide (Fig. 4.18). Compared with WC-Co cemented carbides, they are characterized by improved high-temperature properties. This is especially the case for hot hardness and high temperature strength, oxidation resistance and diffusion resistance to iron materials. Their main area of application is in machining long-chipping steels (Fig. 4.4). Due to the mixed carbide content, the cemented carbides of this group can be classified according to their use into two subgroups.

Group A: Mixed carbide content >10–11 mass %. Due to their mixed carbide content, the cemented carbides of this group are characterized by a high level of

Cemented carbide Application group (DIN ISO 513)	HW - P10	HW - P15	HW - P25	HW - P30	HW - M10	HW - M15
Composition (mass percentage)						
WC	60.0	64.5	72.7	78.5	84.5	82.5
(Ti, Ta, Nb)C	31.0	25.5	17.3	10.0	9.5	11.0
Co	9.0	10.0	10.0	11.5	6.0	6.5
Density (g/cm ³) (ISO 3369)	10.6	11.7	12.6	13.0	13.1	13.3
Hardness HV 30 (ISO 3878)	1560	1500	1490	1380	1700	1550
Pressure resistance (cylinder test) (N/mm ²) (ISO 4506)	4500	5200	4600	4450	5950	5500
Bending strength (N/mm ²) (ISO 3327)	1700	2000	2200	2250	1750	1900
Young's modulus (10 ³ N/mm ²) (ISO 3312)	520	500	550	560	580	570
Fracture toughness (N·m ^{1/2} /mm ²)	8.1	9.5	10.0	10.9	9.0	10.5
Poisson constant	0.22	0.23	0.22	0.23	0.22	0.22
Heat conductivity (W·m ⁻¹ ·K ⁻¹)	25	20	45	60	83	90
Coefficient of thermal expansion (293 K–1073 K) (10 ⁻⁶ ·K ⁻¹)	7.2	7.9	6.7	6.4	6.0	6.0

Fig. 4.18 Composition and attributes of WC-(Ti, Ta, Nb)C-Co cemented carbides, acc. to KOLASKA [Kola92]

high-temperature wear resistance, a low tendency to diffusion with iron materials and low abrasion. They are utilized above all in steel and cast steel cutting, with the exception of rust and acid-resistant steels with austenitic microstructures.

Group B: Mixed carbide content < 10–11 mass %. The cemented carbides assigned to this group have relatively good high-temperature wear resistance and abrasion resistance. They are especially suitable for cutting rust, acid and heat-resistant steels with austenitic structures as well as for alloyed or hard austenitic/ferritic cast iron materials.

4.3.5.3 TiC/TiN-Co,Ni Cemented Carbides (Cermets)

Cemented carbides based on titanium carbide and titanium nitride with a Ni,Co binder phase are designated as cermets (formed from *ceramic + metal*). Present-day cermets are complex multi-material systems that can contain a number of additional elements such as tungsten, tantalum, niobium, molybdenum or complex carbides, from which intermetallic phases are formed during sintering.

Cermets Application group (DIN ISO 513)	HT - P05	HT - P10	HT - P20
Composition (mass percentage)			
Carbon nitrides	89.0	85.7	82.3
Additive nitrides	0.6	0.8	1.0
Co/Ni	10.4	13.5	16.7
Density (g/cm ⁻³) (ISO 3369)	6.1	7.0	7.0
Hardness HV 30 (ISO 3878)	1650	1600	1450
Pressure resistance (cylinder test) (N/mm ²) (ISO 4506)	5000	4700	4600
Bending strength (N/mm ²) (ISO 3327)	2000	2300	2500
Young's modulus (10 ³ N/mm ²) (ISO 3312)	460	450	440
Fracture toughness (N·m ^{1/2} /mm ²)	7.2	7.9	10.0
Poisson constant	0.21	0.22	0.21
Heat conductivity (W·m ⁻¹ ·K ⁻¹)	9.8	11.0	15.7
Coefficient of thermal expansion (293 K–1073 K) (10 ⁻⁶ K ⁻¹)	9.5	9.4	9.1

Fig. 4.19 Composition and attributes of cermets, acc. to KOLASKA [[Kola92](#)]

The composition and some characteristic attributes of three typical cermet types is given in Fig. 4.19 [[Kola92](#)]. Compared to conventional cemented carbides, cermets are characterized by their lower density. Some essential differences to WC-based cemented carbides are their significantly smaller thermal conductivity coupled with their greater thermal expansion. Due to their lower heat conductivity, a larger portion of the cutting heat is dissipated with the chip, resulting in a lower total heating of the cutting insert. Near the contact zones however, the temperature gradient increases inside the cutting tool material. In association with the large thermal expansion coefficient, this leads to high tensile and pressure stresses in the cutting tool material. The result of this is that cermets react much more sensitively to temperature change than conventional cemented carbides. They thus have a strong tendency to form comb cracks, especially in interrupted-cut machining [[Köni90a](#), [Köni93](#), [Kloc96](#)].

Cermets are very hard, have a low tendency to diffusion and adhesion as well as a high level of high-temperature wear resistance. Due to their high edge strength, high resistance against abrasive wear and low adhesion, cermets are especially good for planing steels. Until now, cermets have been predominantly utilized to machine

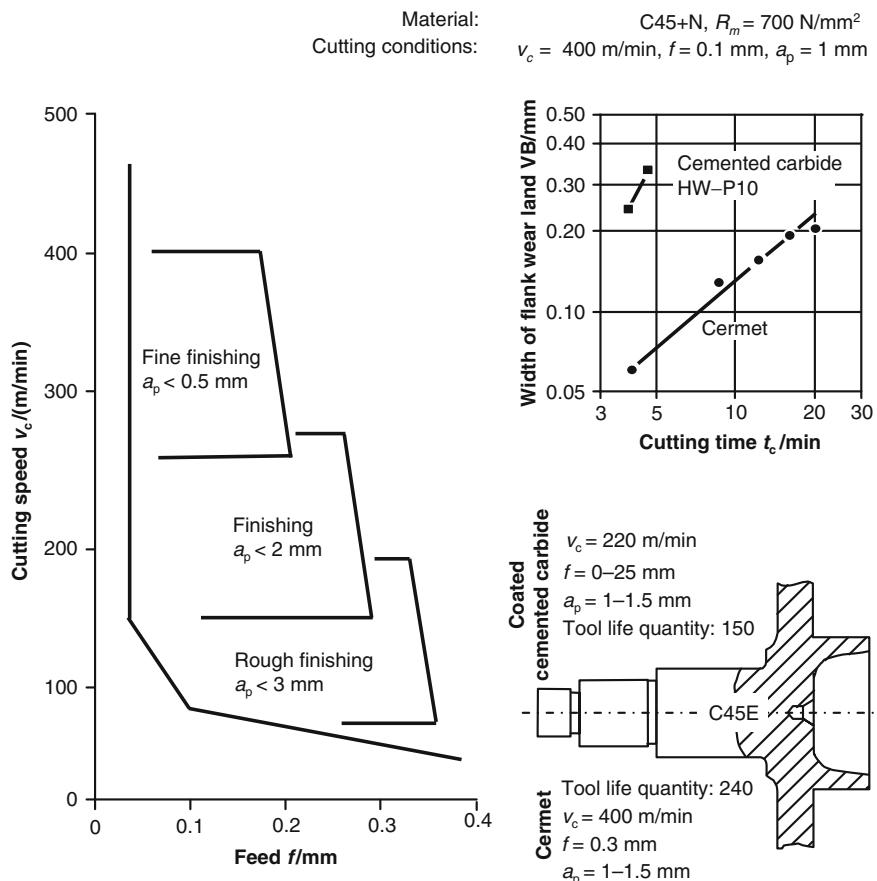


Fig. 4.20 Field of application of cermets during turning steel (Source: Kennametal Widia, CeramTec)

steels with high cutting speeds and small cross-sections of undeformed chip. The development of tougher cermet types has led to the expansion of their field of application to include more average roughing conditions (Fig. 4.20).

Basic areas of application include both turning and milling. Cermets also are suitable for grooving and thread turning. The high wear resistance of the cutting edges, in association with the low diffusion tendency and high oxidation resistance, generally leads to better surface qualities than coated cemented carbides are capable of producing in planing and finishing. At the same time, due to these properties, they allow for higher cutting speeds in comparison with conventional cemented carbides. But they are also suitable for cutting speeds below 100 m/min. Tougher cermet types, corresponding to the range P15 to P25 of conventional cemented carbides based on WC-(Ti,Ta,Nb)C-Co, are used successfully in average rough turning operations and in milling (Fig. 4.21). Because of their clear tendency to

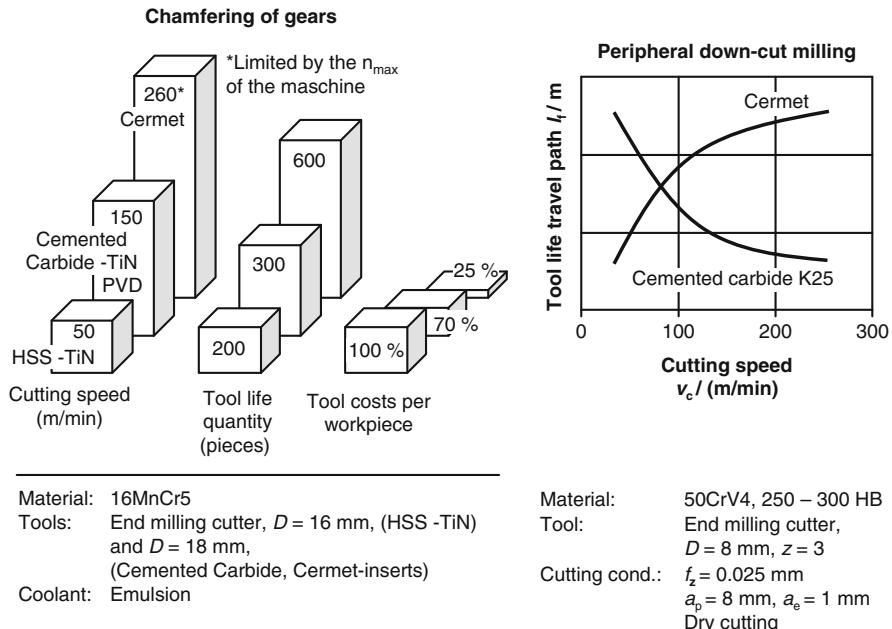


Fig. 4.21 Milling with cermets (Source: VW-Kassel, Kennametal)

form comb-type fractures, cermets should always be dry-milled [Drey97, Köni90a, Köni93, Kloc96].

4.3.6 Uncoated Cemented Carbides and Cermets

Uncoated conventional cemented carbides are still used consistently in cutting technology whenever there are high demands placed in cutting edge sharpness and toughness properties, e.g. in steel milling, finishing, grooving and parting off operations or in thread manufacture. They are however in direct competition with coated cemented carbides and cermets, which are replacing them in an increasing amount of applications. According to DIN ISO 513, uncoated cemented carbides based on tungsten carbide take the abbreviation HW and uncoated cermets the prefix HT (Fig. 4.3).

4.3.7 Function Gradient Cemented Carbides

Function gradient cemented carbides are defined as materials, whose composition and/or microstructure differ in the rim zone and kernel area. The transition from the rim zone to the kernel area is continuous [Berg03].

The goal in the development of function gradient cemented carbides was, and still is, to improve even further the wear resistance of coated indexable inserts (especially in the case of dynamic stress) by modifying the rim zone. Such modifications, e.g. via cobalt enrichening and mixed carbide depletion in a approx. 50 µm thick rim zone, both cracking and fracture sensitivity of the insert are reduced [Berg97]. Such measures are, however, only practical when applied to indexable inserts intended for larger cross-sections of undeformed chip and interrupted cuts. In planing operations, the strength of cemented carbides is generally completely sufficient. In this case, a rim zone that takes over or supports the wear-reducing effect of the coating (mixed carbide enrichening) is advantageous [Berg97]. In the following, three types of function gradient cemented carbides will be introduced.

4.3.7.1 Cemented Carbides with Rim Zone Free of Mixed Carbides

Gradient cemented carbides are understood as cemented carbides, the rim zone of which is free of hard and brittle (Ti, Ta, Nb)C mixed crystals (Fig. 4.22) and consists practically of nothing but tungsten carbide and cobalt up to a depth of about 50 µm. The cobalt content in the rim zone is higher than within. To reduce impact sensibility, a coarse WC grain is advantageous. Moreover, nitrides, which are less hard but more wear-resistant than cubic carbides or carbides of other transition metals (e.g. zirconium carbides) can also be contained in these rim zones. In this way, cemented carbide substrates with high hardness but very tough rim zones of lesser hardness and increased resistance against cracking can be produced.

The fabrication of rim zones free of mixed carbides can take place in various ways. The basic principle behind it is the formation of a nitrogen gradient in the rim zone of the cemented carbides [Berg97].

In the most commonly used method, nitrogen is introduced into the initial cemented carbide mixture as TiN or Ti(C,N). The titanium nitride decomposes during the sintering process. The nitrogen dissolving in the liquid phase diffuses from the cutting tool material into the surrounding atmosphere. In this way, a nitrogen concentration gradient arises in the cemented carbide plate from the inside outwards. The result of this is the formation of a titanium activity gradient. Titanium diffuses into the interior and accumulates there on existing cubic mixed crystals. The same is valid for tantalum and niobium. The mixed carbides near the surface gradually dissolve. On the other hand, cobalt penetrates from inside towards the outside. The growth rate follows a parabolic temporal law. It increases with a rising nitrogen content and decreases with a rising titanium content. An edge effect is characteristic of these processes, the end effect of which is that the zone free of mixed carbides is thinner on sharp edges than on even surfaces [Berg97].

Other method variants include nitrogen pressure treatment, in which the cemented carbide is subjected to a nitrogen treatment under increased pressure after sintering, and the creation of zones free of mixed carbides without nitrogen, in which the cemented carbide contains a defined excess of carbon and a certain sintering regime is adhered to during cooling [Berg97, Köni90, Yohe93].

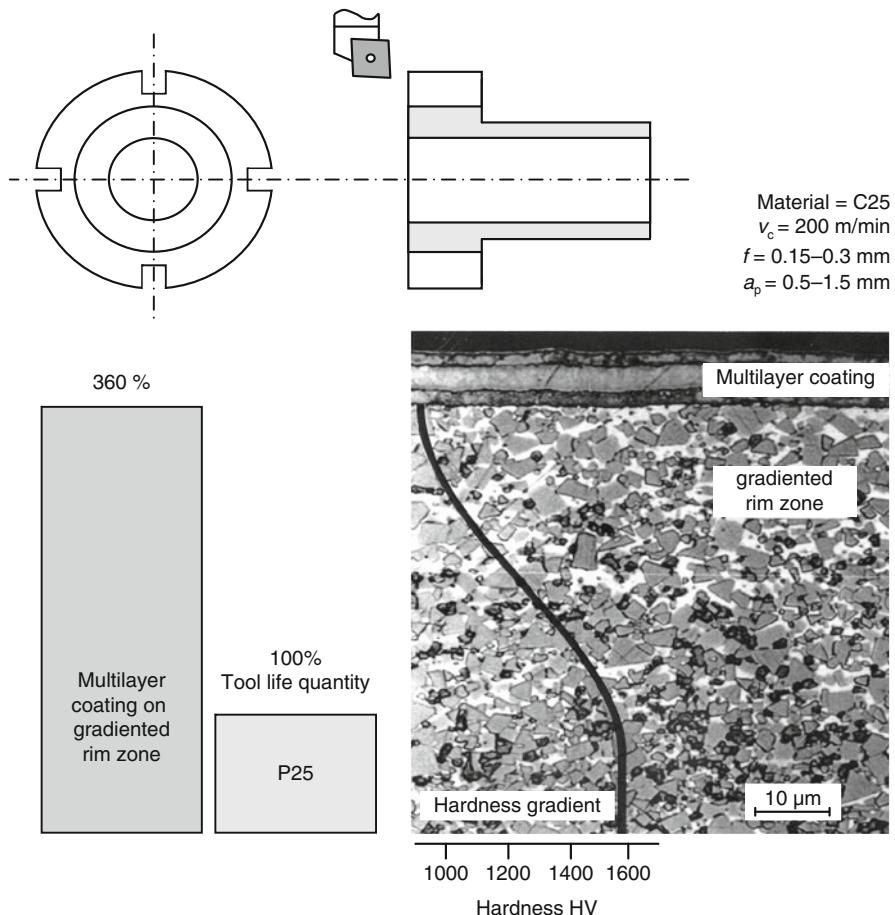


Fig. 4.22 Influence of the rim zone on the tool life of CVD tools (Source: Sumitomo)

The thickness of mixed-carbide-free rim zones produced in this fashion is up to $50 \mu\text{m}$. The methods described can therefore only be used for indexable inserts that are already in their final shape after the sintering process. At best a minor cutting edge rounding can still be undertaken. Therefore, in the case of highly accurate milling inserts that are ground, such toughness-increasing measures are not used, although they would be highly desirable. In the case of turning operations with tools of the application groups P20–P40, indexable inserts with rim zones free of mixed carbides are widely used [Berg97].

As has been shown in practice, thus designed cemented carbides have significant advantages in their wear behaviour compared with coated ones. Especially when turning with interrupted cut, tools coated according to this design achieve significantly longer standing times.

4.3.7.2 Gradient Cemented Carbides with External Ti(C,N) Enrichment

The hallmark of these cemented carbides is an enrichment of nitrides and carbon nitrides on the surface. The substrates are first tightly sintered and then impinged with nitrogen. Diffusion of nitrogen into the rim zone leads to the formation and enrichment of carbon nitrides high in nitrogen on the surface, the composition of which becomes more rich in carbon with increasing depth. The thickness of the surface carbon nitride layer can be varied by changing the treatment time. This process of enriching the rim zone with Ti(C,N) is called coating the substrate during the sintering process. The thus “coated” cemented carbides are efficient alternatives to uncoated and conventionally coated tools. Cutting experiments with gradient cemented carbides with Ti(C,N)-rich rim zones showed clearly improved wear behaviour compared with uncoated cemented carbides. They can however also be conventionally coated. This guarantees that the coating adheres better, especially when one succeeds in designing the coating and the rim zone such that neither are clearly delineable from each other after the coating process [Leng06, Leng04].

4.3.7.3 Gradient Cemented Carbides with Internal Ti(C,N) Enrichment

In this substrate variant, first a rim zone free of mixed carbides is produced. Then nitrogen pressure is altered at a relatively high temperature so that nitrogen diffuses into the interior through the mixed-carbide-free WC-Co rim zone. Diffusion of Ti through the WC-Co layer is, however, minimal. The rim zone functions as a membrane that permits the diffusion within of nitrogen but prevents the diffusion without of titanium. By means of this one-sided diffusion of nitrogen, a transition of the WC-Co zone to the base cemented carbide develops, i.e. a zone enriched with Ti(C,N) a few micrometers beneath the surface in the interior of the substrate. In milling experiments, inserts with internal Ti(C,N) enrichment exhibited excellent wear behaviour in comparison with TiAlN-coated standard inserts [Leng06, Leng04].

4.3.8 Coated Cemented Carbides and Cermets

Coating cemented carbides with thin, highly wear-resistant hard material coatings is one of the milestones in their continued development. Coated cemented carbides consist of a relative tough base body (e.g. P20, K20) on which is applied a 5–20 µm thick hard material coating made of carbides (e.g. titanium carbide, TiC), nitrides (e.g. titanium nitride, TiN), carbon nitrides (titanium carbon nitride, Ti(C,N)) and/or oxides (e.g. aluminium oxide, Al₂O₃). Although uncoated cermets are already relatively high in wear resistance, hard material coating also can further improve their wear performance and efficiency just as with the conventional cemented carbides. According to DIN ISO 513, coated cemented carbides and cermets have the prefix HC (e.g. HC-K20).

4.4 Coatings

Since about 1968, cemented carbides have been coated with hard materials in order to improve their wear resistance and thus also their performance capability. Today, the use of coated cemented carbides and high speed steels for machining various materials is state-of-the-art.

The high efficiency of coated cemented carbides is made even more clear by the fact that in the meantime more than 80% of all cemented carbides used in cutting have a highly wear-resistant hard material coating. In turning, the percentage is about 95% and in milling about 60%.

The primary function of the hard material layer is to inhibit contact between the material, thereby reducing tool wear, which is caused by adhesion, abrasion, diffusion and oxidation phenomena (Fig. 4.23). These processes are superimposed to a large extent and are only partially separable in their causes and in their effects on wear. But in general, besides abrasion at low cutting speeds, it is above all adhesion and, at high temperatures, diffusion and oxidation that determine wear on the tool.

With their high hardness and chemical stability, several hard materials possess the required prerequisites for improving resistance effectively against crater, flank face and notch wear at both low and high cutting speeds.

While the hard material layer improves wear resistance, lessens adhesion between the tool and the workpiece and acts as a diffusion barrier, the task of the substrate is to act as an effective support for the hard material layer and to provide the composite body (consisting of substrate and coating) with sufficient hot hardness and, in particular, toughness.

The toughness attributes of a substrate are of decisive importance for applications with interrupted cut. Due to input and output impacts and associated mechanical and thermal stress variations, the toughness of the cutting edge is much more responsible

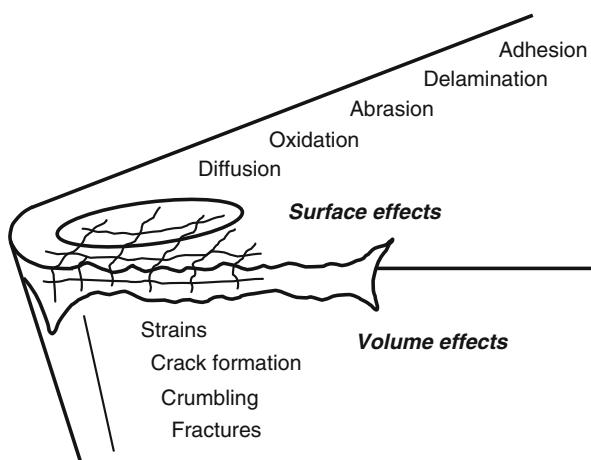


Fig. 4.23 Wear phenomena at coated cutting tools

for its dimensional accuracy than wear resistance. For the interrupted cut, the composite material “coated cemented carbide” must first exhibit sufficient resistance to cracking, crumbling and fracture (Fig. 4.23). Only when these demands are met can, analogously to the interrupted cut, wear mechanisms like adhesion, abrasion, delaminations and – as long as the contact zone temperature and contact time are sufficient – oxidation and diffusion be effectively hindered.

From this we can derive two requirements for coatings [Köni92]:

- The hard material layer has to drastically reduce the effect of all (if possible) wear mechanisms involved in wear process.
- The coating process may not reduce the inner bonding strength, i.e. especially the toughness of the substrate.

4.4.1 Coating Methods

Coating cutting tools can be done chemically and physically. Method variants include

- chemical vapour deposition (CVD)
- and
- physical vapour deposition (PVD).

Figure 4.24 shows an overview of technically established coating methods for the production of wear-resistant hard material layers. As can be seen, layer deposition lies within a specific pressure/temperature range for each method and includes some characteristic coating systems, which are given in bold in the illustration [Berg03].

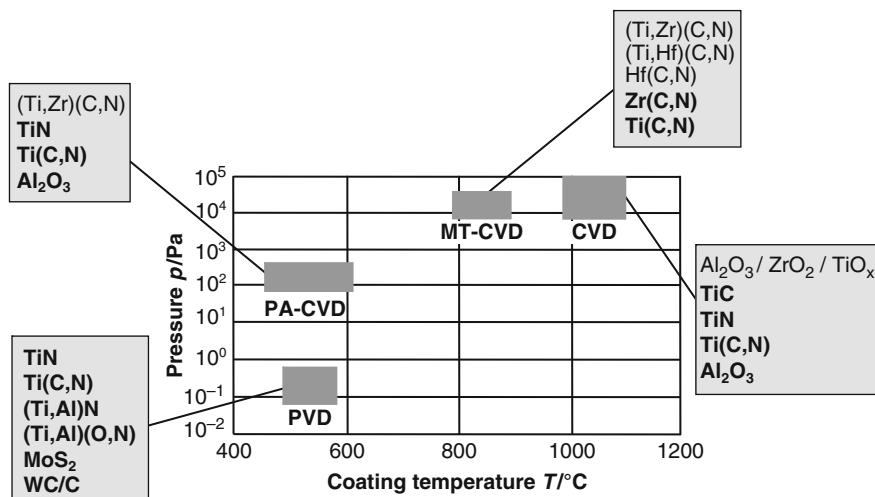


Fig. 4.24 Comparision of pressure and temperature conditions of coating methods with characteristic coating systems (Source: Kennametal Widia)

4.4.1.1 CVD Processes

CVD processes are chemical reactions that take place in gaseous phase under rough vacuum conditions (10^3 – 10^5 Pa) and with the addition of thermal or radiant energy, forming technically useful solids (hard materials) as well as volatile products. The chemical reaction is determined by the laws of thermodynamics and thus by the partial pressure of the gaseous components and temperature.

There are three types of CVD coating procedures:

- HT-CVD (high temperature CVD, 900–1100°C)
- MT-CVD (medium temperature CVD, 700–900°C)
- PA-CVD (Plasma activated CVD, 450–650°C)

A broad palette of hard material coating systems can be synthesized with CVD methods. The most common coating materials are based on hard materials with primarily metallic bonds, such as TiC, Ti(C,N), TiN, heteropolar (ionic) bonds such as Al_2O_3 , but also covalent bonds such as diamond [Damm04]. Figure 4.25 provides an overview of the most common hard materials manufactured with the CVD process.

Figure 4.25 provides a comparison of some essential physical characteristics with the properties of tungsten carbide, the main component of classical cemented carbides.

	WC-6Co	TiC	TiN	ZrC	ZrN	HfC	HfN	$\alpha\text{-Al}_2\text{O}_3$	ZrO ₂	HfO ₂
Crystal system		fcc	fcc	fcc	fcc	fcc	fcc	rhombo-hedral	mono-clinic	mono-clinic
Density (g/cm³)	14.9	4.93	5.44	6.46	7.35	12.30	13.94	3.99	5.7	9.7
Micro hardness HV0,05	1580 HV30	3100	2300	2800	1700	2600	1850	2300	1200	1100
Young's modulus (10³N/mm²)	590	451	570	348	495	352	530	380	200	
Melting point (°C)		3070	2950	3420	2980	3700	3380	2050	2700	
Coefficient of thermal expansion (10⁻⁶.K⁻¹)	5.5	7.7	9.4	6.7	7.3	6.6	6.9	8.0	7.7	5.6
Thermal conductivity (W·cm⁻¹·K⁻¹)	80	33	29	38	21	29	22	6 (at 1000 °C)	0.6 (at 1000 °C)	< 0.1
Colour	Grey	Metallic grey	Gold yellow	Metallic grey	Light yellow	Metallic grey	Green yellow	Uncolored	White	Grey white

Fig. 4.25 Characteristics of hard materials for coatings in comparison to a cemented carbide of the application group HW-K10 (Source: Kennametal Widia)

Because the process is near equilibrium, thermally activated HT and MT-CVD methods can only be used to produce equilibrium phases. On the other hand, the low temperature PA-CVD process is also capable of producing metastable phases such as TiAlN. Among the most technically advanced methods is the deposition of coating systems with nanostructures. In this way, complexly doped hard material phases made of Al_2O_3 or superlattice structures made of TiN/TiB₂ can be produced with a single coating layer thickness of 5 nm [Damm04].

The HT-CVD Process

The high temperature CVD process (HT-CVD) is the classic CVD method for coating cemented carbides. With it, HM tools can be coated at temperatures of 900–1100°C. High temperature coatings are characterized by a high level of adhesive strength on the substrate, which is produced by an easily reproducible material transfer between the surface of the substrate and the reactive periphery [Ever94].

In order to produce a TiC coating, titanium tetrachloride (TiCl_4), for example, is vaporized and transferred with methane (CH_4) to a reaction vessel, which can contain several thousand indexable inserts (Fig. 4.26). Titanium carbide is formed at temperatures of 900–1100°C and a pressure below atmospheric pressure in a hydrogenous environment (H_2) in a chemical reaction following this equation:

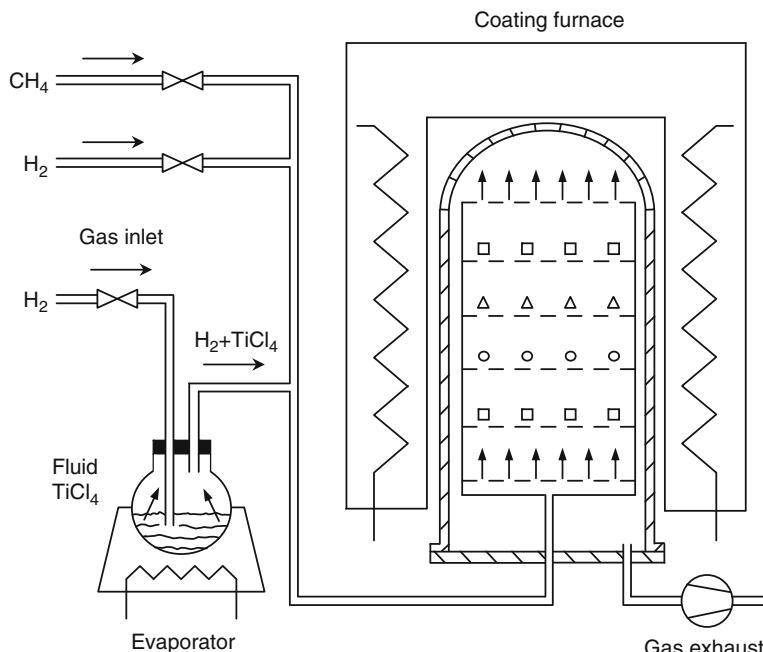
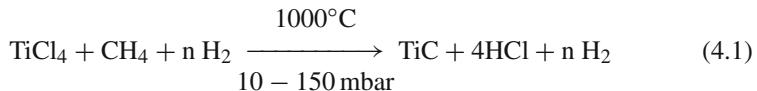


Fig. 4.26 Schematic illustration of a CVD coating unit (schematic)



This process allows us to deposit hard materials like TiC, TiN, $\text{Ti}(\text{C}_x\text{N}_y)$, $(\text{Ti},\text{Hf})(\text{C},\text{N})$, $(\text{Zr}/\text{C},\text{N})$, Al_2O_3 , AlON and others as single coats or in different combinations as multilayer coats. The coating material is formed by chemical reaction from the gaseous phase directly on the surface of the parts to be coated. The reaction products rinse the substrates so that no shading effects arise. Parts with complex geometries can thereby be coated thoroughly and consistently without difficulty.

Cemented carbides that are coated with the classic high temperature CVD process are characterized by high wear resistance due to their relatively thick hard material coatings (up to 20 μm for turning and up to 6 μm in the case of milling).

The high process temperature is problematic in the case of coating tool and heat-treated steels. In this case, coating must be followed by rehardening, making the material vulnerable to unacceptable warpage.

A further disadvantage to the HT-CVD coating process is that the toughness of the coated cemented carbide body is reduced in comparison to the uncoated substrate. The causes for this loss of toughness in HT-CVD coating are extraordinarily complicated. They include on the one hand the properties of the substrate to be coated (chemical composition, crystalline structure, grain size, thermal expansion, bending and pressure strength, pre-treatment) and process control on the other hand (gas atmosphere, pressure conditions, temperature/time cycle). Furthermore, processes taking place during coating in the rim zone (formation of eta phases), residual stresses present in the substrate, transition zone (interface) and hard material layer as well as the microstructure, texture, thickness and adhesion of the applied coating are all mutually responsible for toughness reduction in HT-CVD coating.

We will now consider more closely processes in the rim zone (eta phase formation), the “temperature/time cycle” in the coating process and coating thickness based on all the above factors, which offer us many starting points for the improvement of toughness in coated cemented carbides.

During the high temperature CVD coating process, there is a danger of forming brittle phases in the interface. Cemented carbides of the first generation exhibited an additional approx. 3–5 μm thick brittle zone with an “eta phase” ($\text{W}_6\text{Co}_6\text{C}$, $\text{W}_3\text{Co}_3\text{C}$), caused by decarburization processes in the rim zone [Schi89]. Embrittlement from the eta phase has a negative influence on the toughness of coated cemented carbides.

Figure 4.27 shows the time/temperature transformation chart for a cobalt alloy with 5% tungsten and 0.23% carbon. It shows typical coating temperatures and times for the HT-CVD, MT-CVD, PA-CVD and PVD methods. From this chart, we can see that in the classic HT-CVD and medium temperature CVD process the area of eta phase precipitation is passed through/tangent. In contrast, no changes in constitution are to be expected in the case of low temperature coating methods, as in the PA-CVD or PVD processes [Köni90].

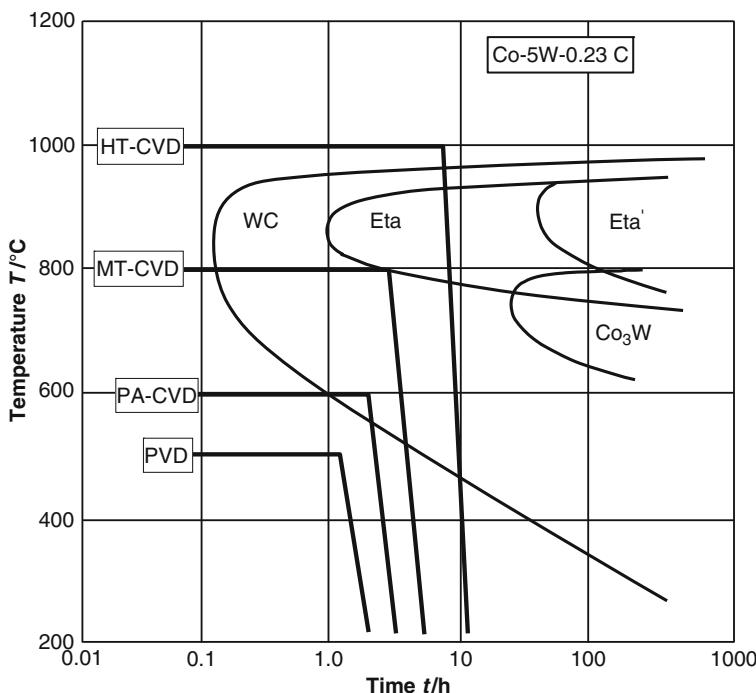


Fig. 4.27 TTT-chart of a cobalt alloy with typical temperature sequence of several coating processes (Source: Widia)

The comparative examination of the bending strength of thin, variously coated WC-Co cemented carbides shows what effect coating temperature and coating thickness have on these toughness-relevant parameters (Fig. 4.28).

With increasing coating thickness and coating temperature, bending strength is reduced. The effect is especially negative in cutting operations with interrupted cut, in which the tool is subject to dynamic stresses. In the case of milling therefore, cemented carbides with thinner hard material coatings (4–6 μm) than in turning are used. With these coating thicknesses, toughness loss is relatively low. Reduction of wear protection due to the reduced coating thickness must be expected.

The logical consequence of these discoveries is that more and more coating methods have gained significance for coating cemented carbides in the past few years, which can be used at lower coating temperatures in comparison with the classic HT-CVD process.

The MT-CVD Process

By using acetonitrile ($\text{CH}_3\text{-CN}$) instead of a methane (CH_4)/nitrogen (N_2) gas mixture, it is possible to lower the coating temperature in comparison with the HT-CVD method by up to 200°C [Chat86]. In the 700–900°C temperature range, carbon

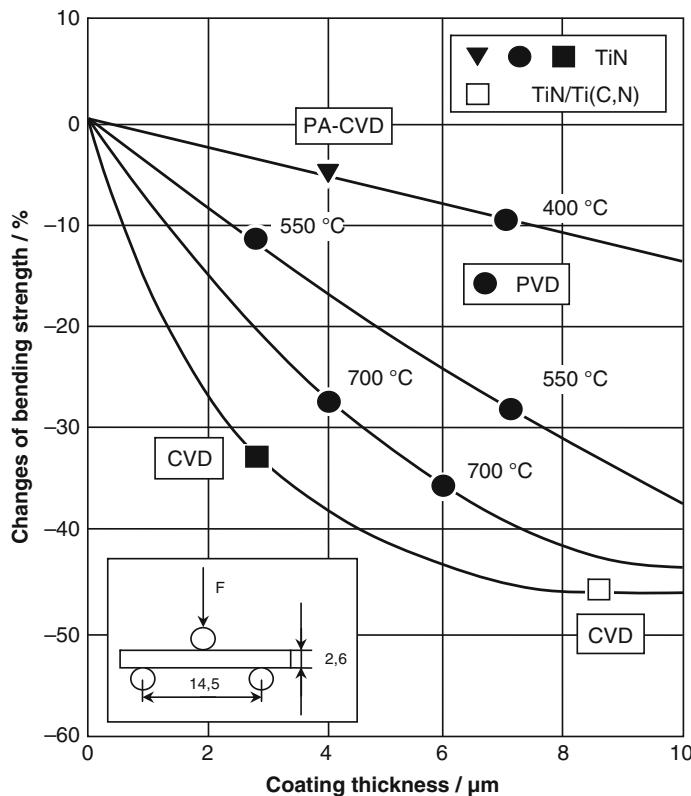


Fig. 4.28 Influence of coating method and coating thickness on the bending strength of coated cemented carbides (Source: Widia)

nitride coatings of the elements Ti, Zr and Hf can be applied with high deposition rates. As opposed to the conventional HT-CVD process, the MT-CVD method has the following advantages:

- Thermal stress on the cutting tool material is lower due to the reduced coating temperature at simultaneously higher deposition rates.
- The danger of decarburization and thus of embrittling eta phases is lower for cemented carbide substrates.
- The coatings applied with the MT-CVD method exhibit lower tensile residual stresses compared with HT-CVD coatings.

Trial results from the longitudinal turning test show the same qualitative dependence on cutting speed for MT-CVD coated cemented carbides as for the uncoated substrate (Fig. 4.29) [Köni92, VDI324]. Both MT-CVD coated and PVD coated cemented carbides are clearly superior to uncoated cemented carbides in performance. In higher cutting speed ranges, the performance of the 3 μm thick

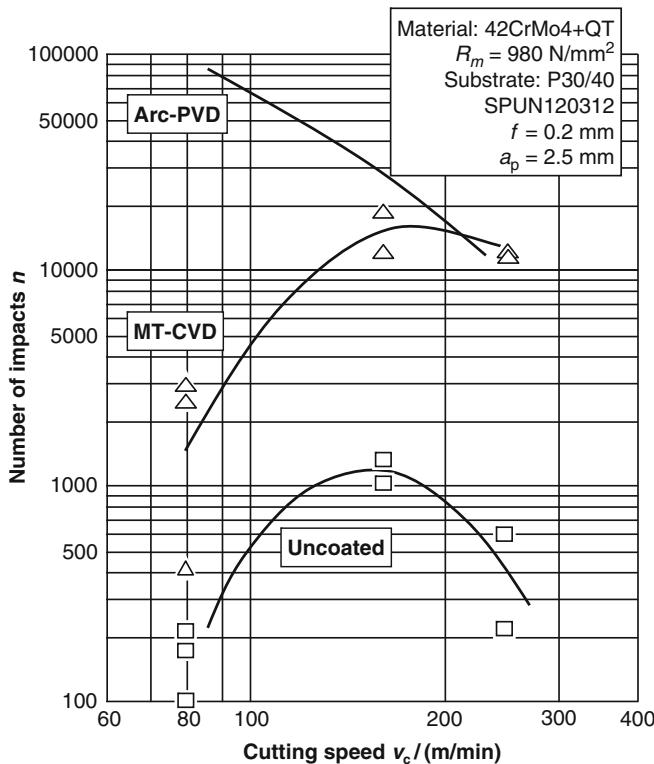


Fig. 4.29 Influence of coating methods and cutting speed on the toughness behaviour of carbides in an interrupted longitudinal turning test

Arc-PVD-TiN coating and the 6 μm thick MT-CVD-TiN-Ti(C,N)-TiN coating do not differ. The PVD coating tends to show advantages at low cutting speeds. It extends the range of applicable cutting speeds clearly to lower cutting speeds to a range that is also covered by coated HSS. At low cutting speeds, the cutting edges succumb to flank face wear and no longer to cracking and crumbling. The causes for the improved toughness are the unaffected properties of the substrate and the more favourable compressive residual stresses resulting from PVD coating of the cemented carbides [Köni90].

MT-CVD coated cemented carbides are used above all in milling. There are significant performance advantages compared with conventional coatings. The end of their standing times is no longer brought about by cutting edge crumbling, but in a reproducible way by a continuous wearing of the flank face. Their excellent toughness properties make these cemented carbides also ideal for milling with cutting fluids. Fewer comb cracks form and crack development is slower than it is the case with HT-CVD coatings. Transverse cracks and cutting tool material crumbling occur later, resulting in longer tool standing times and higher machining quality.

Residual Stresses in CVD Coating Systems

As a rule, coatings deposited by means of the HT or MT-CVD process exhibit tensile residual stresses. This is, however, no longer exclusively true. In the meantime, there are also CVD systems with compressive residual stresses available.

Tensile residual stresses originate when cooling from coating temperature to room temperature and result primarily from the difference of the thermal expansion coefficients between the substrate and the coating or, in the case of multilayers, between the individual coatings. In addition, stress-free annealing in the substrate cemented carbide always occurs at the coating temperatures typical for the CVD process, leading (as opposed to the low temperature PVD process) to a reduction of the (compressive) residual stresses introduced, for example, during grinding [West00].

Coatings with compressive residual stresses, generally familiar in the case of PVD coatings, have the advantage that crack development is hindered in the hard material coating, and the fracture strength of the coating/substrate composite is largely maintained. Compressive residual stresses are therefore the most beneficial to performance in interrupted turning and in milling in general.

Compressive residual stresses can also be produced when using the CVD method by a corresponding choice and arrangement of individual coating systems. One example is the coating sequence TiN-Ti(C,N)-Al₂O₃. The outside coating composed of Al₂O₃ has the smallest thermal expansion coefficient, the inside TiN coating the largest. After cooling from coating to room temperature, the Al₂O₃ coating exhibits a small amount of compressive residual stress. Considerable larger compressive residual stress is measured with the same coating sequence if it is supplemented externally with a further Zr(C,N) or Hf(C,N) coating (ratio C:N = 1:1) (Fig. 4.30).

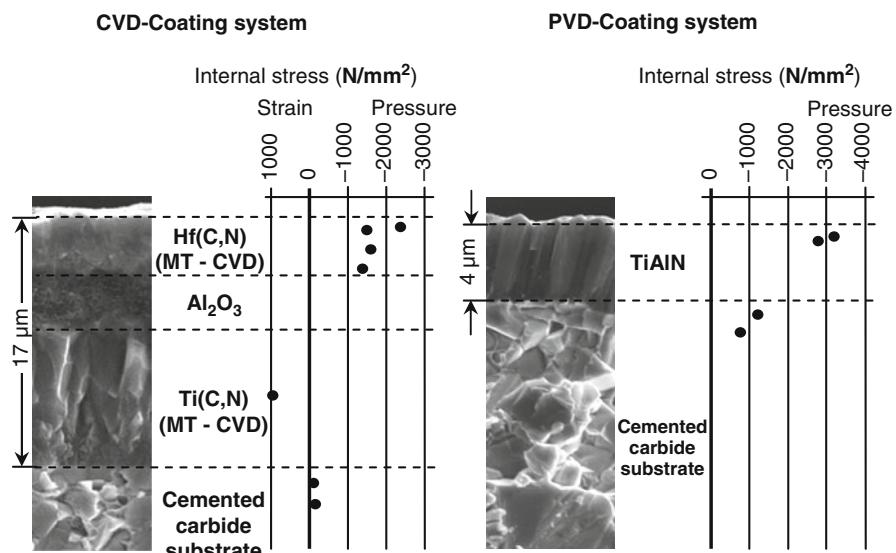


Fig. 4.30 Influence of coating methods and coating structure on residual stresses in hard coating systems

To produce these multilayer systems, the classic high temperature CVD process is combined with the MT-CVD process, whereby the precipitation of the carbon nitride coatings takes place with the help of the medium temperature CVD process [Drey97, West00, Berg03, Berg05].

The Plasma Activated CVD Process

In the PA-CVD process (“plasma assisted” or “plasma activated”) – also known in the literature as PE, “plasma enhanced” – precipitation of fine-grained hard material coatings takes place at the, in contrast to HT and MT-CVD methods, much lower coating temperatures of 450–650°C. At these temperatures, the thermal energy alone is insufficient for introducing the chemical reaction needed to form the hard materials from the gaseous phase. For this reason, additional energy is added to the process by means of the pulsed plasma of a low pressure glow discharge. In this way, chemical reactions are made possible that, at thermodynamic equilibrium, are only possible at much higher temperatures [Tabe89, Köni89].

As in the HT-CVD process, coatings made of titanium nitride, titanium carbide, titanium carbon nitride and aluminium oxide can be deposited individually in crystalline structure by changing the gas composition with individually or in an arbitrary sequence. Due to the non-equilibrium conditions during deposition, there is also a possibility of synthesizing metastable systems like (Ti,Al)N using PA-CVD [Lemm03].

The attributes of the cemented carbide remain largely unaffected by the lowering of temperature during coating. The residual stresses are in the compressive range after coating. PA-CVD coated cemented carbides measure about 30% higher bending strengths than HT-CVD coated ones.

The properties of PA-CVD coated cemented carbides have a positive effect on their performance when cutting strong steel materials in interrupted cut. The sensitivity of the composite to comb cracking and failure due to crumbling is much lower than it is the case for HT-CVD coated cutting tool materials. The application example “side milling” proves this (Fig. 4.31). As opposed to the uncoated substrate, the tool life travel path is slightly diminished after HT-CVD coating due to thermal influence on the substrate as well as tensile residual stresses in the hard material coating, while it is over 300% larger after PA-CVD coating.

Modifying the Layer Structure

The layer structure of many multilayer systems usually begins with a thin titanium nitride layer coating the substrate, followed by one or more titanium carbon nitride intermediate layers. The TiN surface layer functions as a diffusion barrier during coating and prevents brittle eta phases forming in the interface [Köni90]. The Ti(C,N) layers building upon the TiN surface layer can consist of one layer or several layers differing in stoichiometry. They act as a kind of “buffer” between the relatively soft and tough substrate and the much harder and more brittle carbides, nitrides or oxides of the coating.

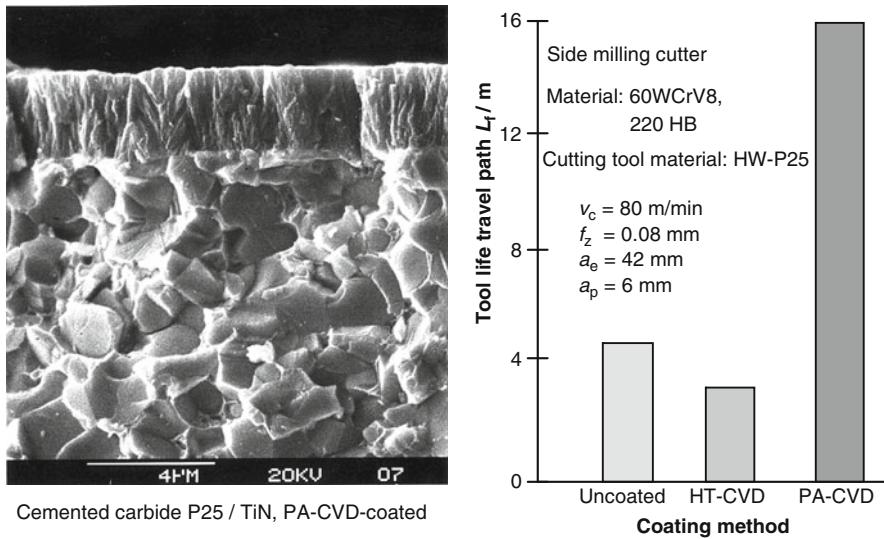


Fig. 4.31 Cutting edge durability of a PA-CVD-coated cemented carbide in side milling (Source: Widia)

With increasing coating thickness, the wear resistance of the CVD-coated cemented carbides increases, but their bending strength and toughness is reduced. The toughness of the coating/substrate composite is affected not only by the thickness of the entire coating, but also to a large extent by the coating material and the thickness of the individual coating components.

Figure 4.32 illustrates this using the example of a coating made of TiC and Al₂O₃. With a constant total coating thickness of 10 μm, the failure rate of the coated cutting tool material due to fracture clearly increases with the thickness of the Al₂O₃ layer [Naka88]. Due to its high chemical stability as a coating material, aluminium oxide does indeed have excellent resistance against crater wear, but it only has a small amount of thermal shock resistance and toughness.

In order to reduce the brittleness of coating systems containing aluminium oxide, they are often designed in the form of fine lamellar, fine grained multilayer coatings (Fig. 4.32). Such multilayer coatings consist usually of a combination of TiN or Ti(C,N) with Al₂O₃ or, as in Fig. 4.32, with aluminium oxynitride (AlON). Coatings such as these can also be composed of a large number of individual layers, whereby each layer can be thinner than 0.2 μm. Total coating thicknesses, depending on the intended use, are around 5 μm in the case of milling and between 12–20 μm for turning.

Using PA-CVD coating technology, we are now also capable of depositing nanolayer coatings. These are multilayer coatings with individual layers whose thickness is given in nanometres, not in micrometers (Fig. 4.33). In the case of coating systems 1 and 2 shown in Fig. 4.33, sixty 80 nm thick layers and thirty-six 50 nm thick layers are deposited on a top layer composed of titanium nitride. In

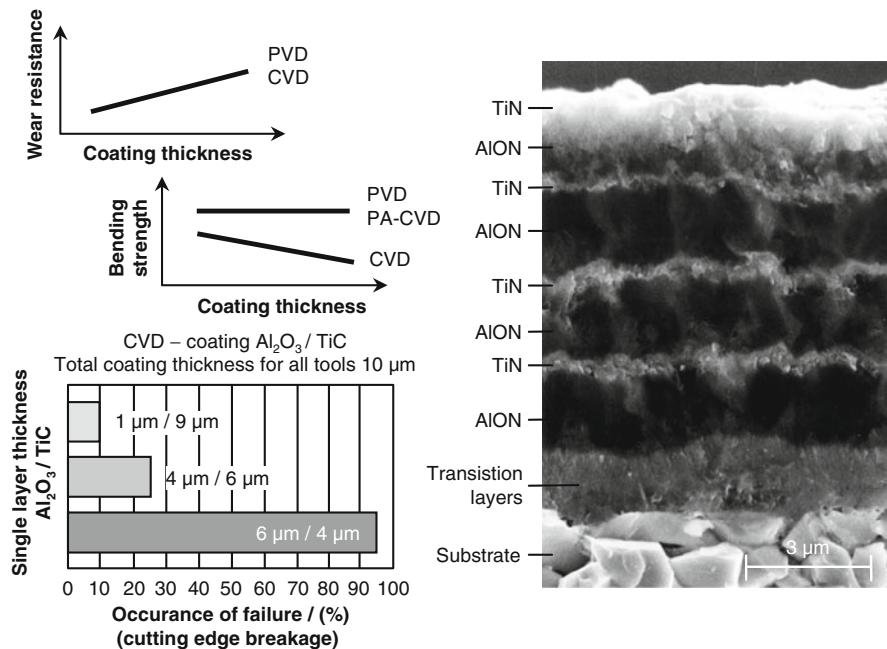


Fig. 4.32 Fine lamellar multilayer coating (Source: Widia, Sumitomo)

the case of the coating system TiN-AlN, there are as much as 2000 single layers of titanium and aluminium nitride of 9 nm thickness that are applied alternately.

Lowering the thickness of the coating increases hardness, toughness and wear resistance. This effect is based, among other things, on the fact that coating materials below a certain thickness take on completely new properties. For example, AlN is, with a hardness of about 1200 HV0.05, relatively soft in comparison to hard materials like TiC, however it has a hardness of over 3000 HV0.05 when the coating thickness is below 10 nm.

At high cutting speeds or during dry machining, demands on thermal resistance of the tool cutting edge are increased. Due to their low thermal conductivity, oxide coatings are especially suitable to meet these demands. As can be seen in Fig. 4.25, in addition to aluminium oxide, both zirconium oxide ZrO_2 and hafnium oxide HfO_2 have very low thermal conductivity. Since both oxides are insufficiently hard by themselves, they are combined with Al_2O_3 [West00]. This can be accomplished in the form of multilayer coating systems with single or multi-phase layers, whereby the latter – so-called composite coatings – are finely distributed in the Al_2O_3 base matrix of a further oxide, e.g. HfO_2 oder ZrO_2 (Fig. 4.34).

In the case of the $\text{Al}_2\text{O}_3/\text{ZrO}_2/\text{TiO}_x$ composite coatings shown in Fig. 4.34, the oxide layer, consisting of three components, is condensed from a gas mixture comprising AlCl_3 , ZrCl_4 , TiCl_4 , CO_2 , H_2 , N_2 and Ar. Al_2O_3 then exists as a κ -phase and ZrO_2 as a phase mixture in tetragonal/monoclinical (coating temperature > 990°C)

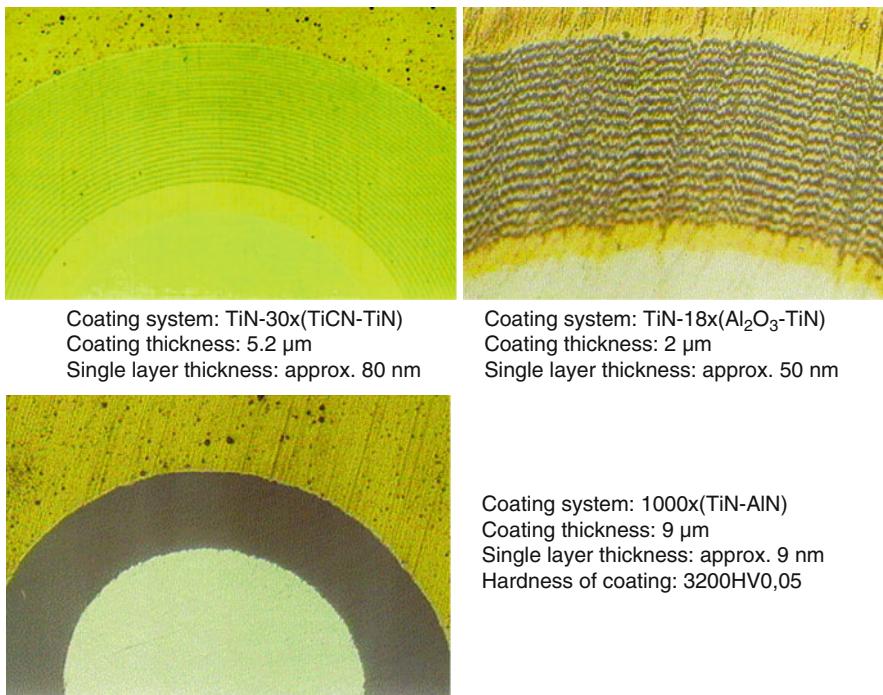


Fig. 4.33 Nanolayer created by a PA-CVD coating process (Source: Kennametal Widia)

or monoclinical form (coating temperature < 990°C) depending on the coating temperature. In the fracture structure image, the ZrO₂ particles can be recognized as finely distributed in the Al₂O₃ matrix. The highly fine-grained coating structure is originated by means of a small addition of titanium as doping agent.

Doping layers made of Al₂O₃ with titanium or boron (or TiN layers with boron) leads to the formation of very fine-grained coating structures. In the CVD process, titanium is added to the gas mixture as TiCl₄ and boron as BCl₃. The doping products TiO_x und B_xO_y, well detectable in the EDX (energy dispersive X-ray) analysis but still not clearly identifiable as a phase, are effectively dissolved in the Al₂O₃ lattice. Due to their nanodispersive distribution, they have a positive effect on coating growth and the grain size of the hard materials forming the coating system [West01, Kloc01, Berg03]. Doping brings about higher deposition rates, makes it possible to reduce deposition temperature and clearly improves wear resistance at high thermal cutting edge loads (Fig. 4.35) [Kath03].

Aftertreatment

Aftertreatment helps to improve further the performance characteristics of coatings, especially in the case of interrupted cut. For example, smoothing the surface of CVD-coated tool bodies by means of brushing, jet treatment or polishing is

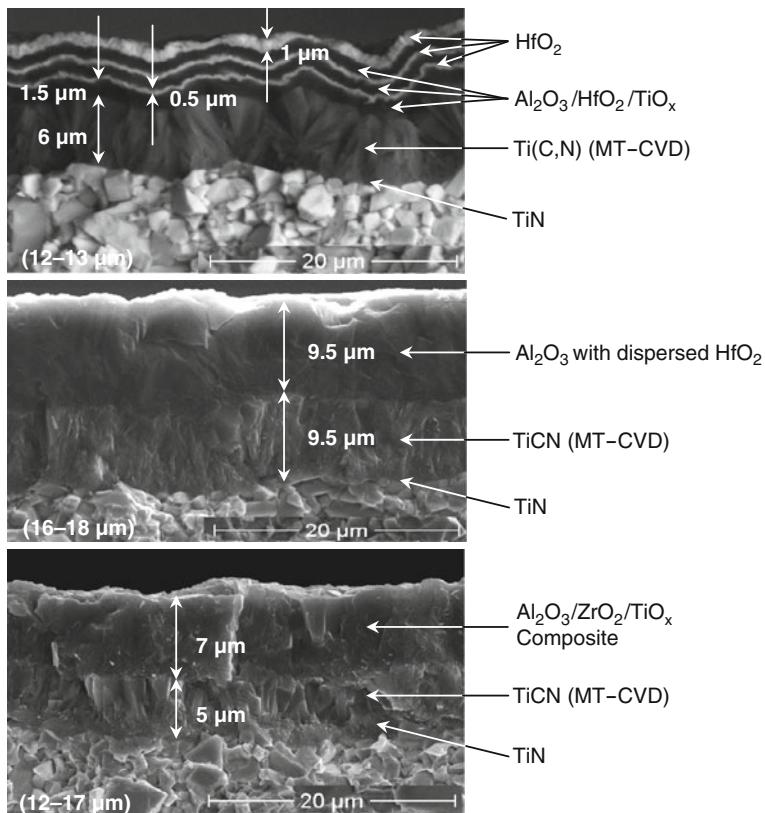


Fig. 4.34 Fracture structure of CVD-coated cemented carbides with modern coating-design
(Source: Kennametal Widia)

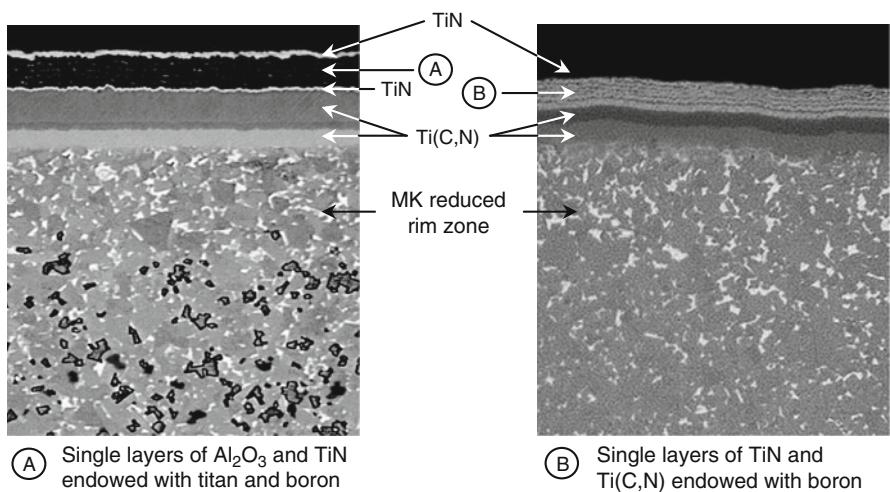


Fig. 4.35 Formation of microcrystalline layer structures via doping with titanium and boron
(Source: Ceratizit)

the common practice today for removing potential cracking sources. Since brushing only covers the cutting edge area, all-round smoothing with jet treatment is being given increasingly more preference. It is of particular importance that the jet treatment, in addition smoothing, also causes an alteration of the tensile residual stresses as a function of the treatment parameters. For example, the Al_2O_3 layer in the $\text{Al}_2\text{O}_3/\text{ZrO}_2/\text{TiO}_x$ coating system has, after jet treatment, compressive residual stresses (-3.500 N/mm^2) in place of tensile residual stresses (1000 N/mm^2). Jet treatment not only improves the residual stress condition, but also reduces the range within which the wear and standing time of coated hard material indexable inserts are scattered [Berg03].

Modifying the Substrate Rim Zone

In comparison to uncoated cemented carbides, CVD coatings lengthen standing times enormously. On the other hand, they also lower the toughness of indexable inserts, thereby also reducing their capacity to bear high stress peaks during cut interruptions [Köni90, Köni92]. The reason for this is assumed to be cracks, which spread from the hard tensile stressed material coating into the cemented carbide substrate.

Attempts to improve the toughness and resistance of cemented carbides substrates to crack development without reducing their hot hardness led to the development of substrates with graded rim zone structures [Sche88, Köni90, Schw89]. The goal was to modify the rim zone of the cemented carbide such that crack development towards the centre of the indexable insert was hindered or suppressed. One possible solution are cemented carbides, the rim zones of which are largely lacking hard and brittle $(\text{Ti}, \text{Ta}, \text{Nb})\text{C}$ mixed crystals and consist practically of only tungsten carbide and cobalt up to a depth of approx. $50 \mu\text{m}$ (mixed carbide-free or reduced rimzone, Fig. 4.35). The coating and application of such cemented carbides is presently state of the art.

Further developments in this area are aimed at establishing the basic properties of a coated cemented carbide body – hard outside and tough inside – already during sintering [Leng04]. This would also have positive effects on the following coating process. The fabrication of function-graded cemented carbides with rim zones lacking mixed carbides and with external or internal $\text{Ti}(\text{C},\text{N})$ enrichment is described in more detail in Sect. 4.3.5.

4.4.1.2 PVD Coating

The early 1980s saw the succession of the CVD process by the PVD process in various process variants (vacuum evaporation, Arc-PVD, sputtering). The initial benefit of these methods was the possibility of coating HSS tools with complex geometries. In the meantime, these processes are also being used for cemented carbide and cermet coating. In addition, decorative layers, oxidation layers and corrosion layers can also be produced.

The fundamental differences between PVD coating and the classic HT-CVD process can be summarized as follows [Sche88, Schi89, Köni90, Rödh87, Knot89, Gühr89, VDI3198, VDI3824]:

- The process temperature of 160–600°C means relatively low temperature stress on the substrate materials, so that temperature-sensitive substrate materials can also be coated.
- The bending strength of the substrate remains largely unaffected by the low coating temperature.
- PVD coatings exhibit residual compressive stresses, which currently limit the possible layer thicknesses to 3–6 µm. Residual compressive stresses reduce the risk of fracture formation under shock load and thermal cycling.
- In the PVD process, workpieces that are to be coated require a very careful pretreatment of the surface and process control in order to guarantee sufficient adhesion of the coating. CVD coatings have better adhesion due to the effects of evaporation and diffusion.
- Because of shadowing effects, consistent layer thicknesses can only be partially realized by means of very costly devices for rotating the parts to be coated. Inner contours are as a rule only coatable up to a depth/diameter ratio of 1, since the layer thickness is diminished with increasing depth.
- High number of potential coating systems and substrate materials.
- Besides stable coating systems such as TiN and Ti(C,N), multi-component metastable phases can also be deposited under the non-equilibrium conditions of physical vapour deposition. These phases often exhibit structures that do not exist under equilibrium conditions. An example of this is the coating system TiAlN.

We can distinguish several different methods of PVD coating:

- methods with thermal evaporation (vacuum evaporation)
- methods with arc evaporation (arc-PVD)
- methods with cathode evaporation (sputtering)

Thermal Evaporation (Vacuum Evaporation)

In vacuum evaporation, usually the coating material is evaporated in a crucible with a resistance heater or an electron beam gun (EB-PVD) in high vacuum at a pressure of 10^{-3} – 10^{-6} Pa (Fig. 4.36). The vapour atoms exhibit average free path lengths of up to several meters at these pressures. Therefore, they generally do not interact with each other and proceed straight to the substrate. Since the latter is considerably colder than the vapour, the particles condense on the substrate in accordance with the principle of water vapour. A columnar fine crystalline layer is formed [Möh96]. Due to the straight path of the particles, the substrate material must be moved in the coating chamber in order to avoid shadowing effects and inconsistent coating thicknesses. Vacuum evaporation allows for high coating rates (coating thickness increase per unit time) and is used above all for optical applications, in mass production for large-scale metallization of films, plastic parts and paper as well as for coating turbine blades (EB-PVD) [Sche88, Rass96].

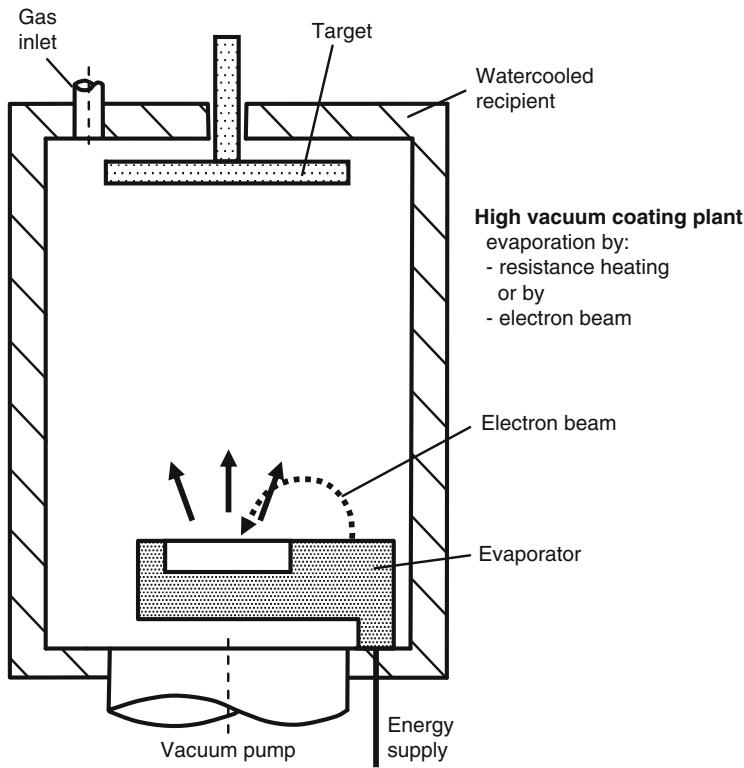


Fig. 4.36 PVD-Process – Vacuum evaporation, acc. to Möhwald [Möhw96]

In order to form hard material coatings composed of carbides, nitrides or oxides, the substrate is reactively vapour-deposited with the help of a reaction gas (N_2 , CH_4 , ...). [Haef87]. For example, from $2Ti + C_2H_2$ the hard material layer $2TiC$ and the gas H_2 are formed. In the case of simple reactive evaporation, bonding strength is relatively low due to the low particle energy of less than 2 eV.

Arc Evaporation (Arc-PVD)

In the arc-PVD method, the target material is vapour-deposited and ionised by means of an arc in high vacuum. The arc travels either randomly (random arc) or in a controlled fashion (steered arc) by means of the vaporizing source, usually connected as a cathode (Fig. 4.36). This method is distinguished by its high degree of ionisation of up to 90% and thus makes it possible to produce high quality coatings with thick structures and good adhesive properties.

Due to the high energy density in the focal point of the arc, the material is explosively evaporated and propelled, but molten particles are also torn out at the same time. The liquid particles that are carried away are found again as drop-shaped defects (“droplets”) in the layer [Bran05]. These are undesirable because they adhere to the substrate insufficiently [Möhw96]. The appearance of droplets is

the main disadvantage of arc evaporation, and since the droplets can lead to pores in the coating, this method is not suited to depositing corrosion-resistant coatings [Möhw96]. Droplet emission can be partially reduced by optimizing the process parameters. Since evaporation by means of cathode arcs requires that the coating material is electrically conductive, no oxide coatings can be deposited with this method.

One process variant is the LARC method (Lateral Rotating Arc Cathodes). The basis of this method are rotating, water-cooled arc cathodes [Damm04]. The magnetic field is created by permanent magnets and coils that are shifted and steered vertically and radially. In contrast to conventional arc technology, fewer droplets are produced, resulting in smoother coatings [Csel03].

Cathode Evaporation (Sputtering)

In a low-pressure plasma, an inert gas (e.g. argon) is ionized by applying high voltage. The positively charged inert gas ions are propelled onto the target switched as a cathode (coating material) and knock out atoms, atomic groups and molecules of the coating material via impulse exchange [Rass96] (Fig. 4.37).

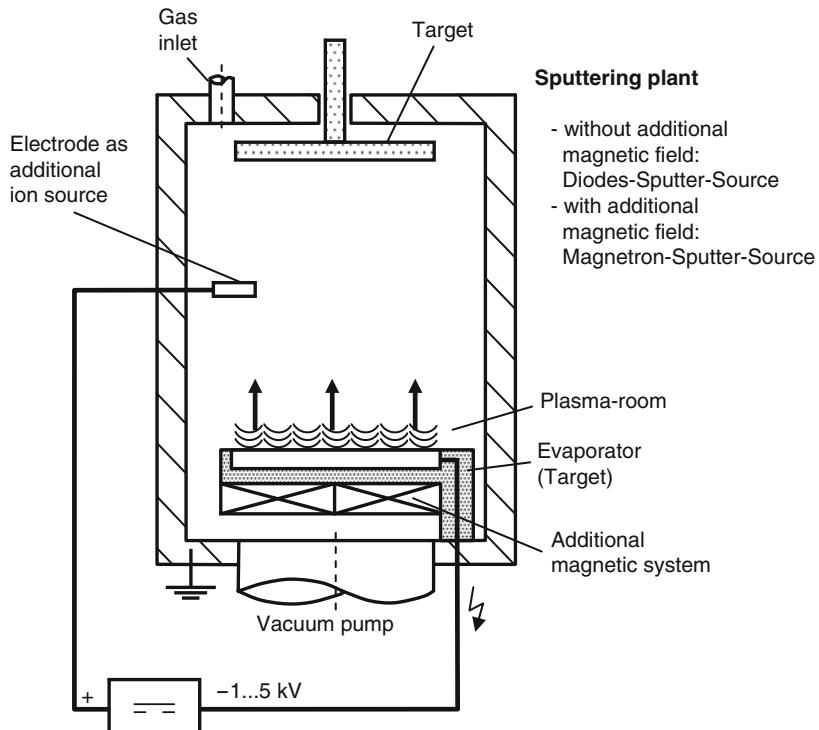


Fig. 4.37 PVD-Process – Cathode evaporation (sputtering)

As opposed to vacuum evaporation, sputtering requires a process gas (usually argon) as well as a high voltage source. After evacuating the recipient, the process gas is admitted at up to 0.1–1 Pa pressure. The average free path length of the particles at this pressure amounts to a few millimetres. At a target voltage of –1 to –5 kV, a glow discharge plasma is created. The plasma burns between the target which acts as an anode and the substrate connected to the ground or the recipient wall, thus making it the anode. The plasma consists of ions, electrons and uncharged atoms [Möhw96].

This evaporation method is universally applicable, since the material is evaporated not thermally but by means of impulse transfer. Almost all substances [Haef87] and composites of very diverse melting substances can be evaporated, i.e. not only metallic, but also isolating layers as well as materials with high melting points can be deposited.

If not only metals or metal alloys are to be deposited, but also carbides, nitrides, or oxides, this can be achieved with reactive sputtering. In this case, a reaction gas (e.g. N₂, CH₄, O₂, H₂, ...) is injected in addition to the process gas. From this reaction gas are obtained the non-metallic components of the coating (e.g. TiN, TaC, Al₂O₃, ...).

As in the arc method, sputtering also involves the possibility of depositing metastable coatings (i.e. coatings that do not exist in equilibrium). Because the process conditions are far from equilibrium conditions, it is possible to freeze metastable composite phases due to their rapid condensation from the gaseous phase, thus synthesizing a completely new material type that is often superior to stable boundary systems [Leye04].

We differentiate between DC sputtering, in which the sputtering material is electrically conductive, and RF sputtering, in which there is a high-frequency AC field that is used for non-conductive materials. The use of direct current (DC) plasma, high-frequency (RF) plasma and the combined use thereof as well as a nearly unlimited number of possible coating materials make sputtering enormously flexible with respect to the development of application-specific component coatings [Rass96]. Although RF sputtering has important advantages when atomizing isolating materials, this technology is of lesser importance in the creation of wear-protection coatings due to the appreciably higher capital costs in comparison with DC sputtering and the significantly lower deposition rates [Lemm03]. The disadvantages of RF sputtering can be surmounted by using pulsed DC sputtering (MF: medium frequency sputtering, up to 50 kHz). The coatings produced with sputtering at low vapour-deposition rates have good adhesive properties.

Magnetron Sputtering

In comparison to pure diode sputtering, magnetron sputtering involves the arrangement of permanent magnets behind the target (Fig. 4.37), the magnetic field lines of which vertically penetrate the cathode plate and the target affixed onto it. Due to the overlapping of magnetic and electric fields, the electrons in the plasma are

directed on extended spiral-shaped paths. In this way, the probability of ionisation is increased as well as the deposition rate as the discharge voltage is reduced. It was the invention of magnetron sputtering that has first made possible an economic utilization of the sputtering effect for coating deposition and thus the commercial deposition of wear-protective layers by means of sputtering technology [Lemm03, Bobz00].

High-Ionisation Pulsing

Rising interest in isolating, mostly oxidic wear-protection coatings and the necessity of increasing ion density in the area of the substrate in order to deposit layers of dense, optimized morphology have led to the development of pulsed DC plasmas. With this technology, not only coating systems usually reserved for RF technology can be deposited, but completely new coating systems can also be synthesized due to the extremely enhanced ion currents [Lemm03, Leye04].

Pulse technology makes it possible to deposit conductive and electrically isolating layers in almost arbitrary stoichiometry as well as nano-structured single and multi-phased layers with new property combinations.

Figure 4.38 shows a (Ti,Al)N coating created with conventional DC technology in comparison to one made with the help of pulse technology. One can clearly recognize the alteration of the coating micromorphology towards finer crystalline structure made possible with the increased plasma ionisation of pulse technology [Lemm03]. Pulse technology thus also creates the possibility of depositing crystalline γ -Al₂O₃ at substrate temperatures of < 500°C or producing (Ti,Al)N layers with a AlN content of up to 66 mol% (Fig. 4.39).

The coating process with pulsed plasmas is also known as high ionisation pulsing (H.I.PTM). We distinguish between three process variants: unipolar, asymmetrical bipolar and bipolar two-cathode magnetron sputtering [Lemm03, Erke05].

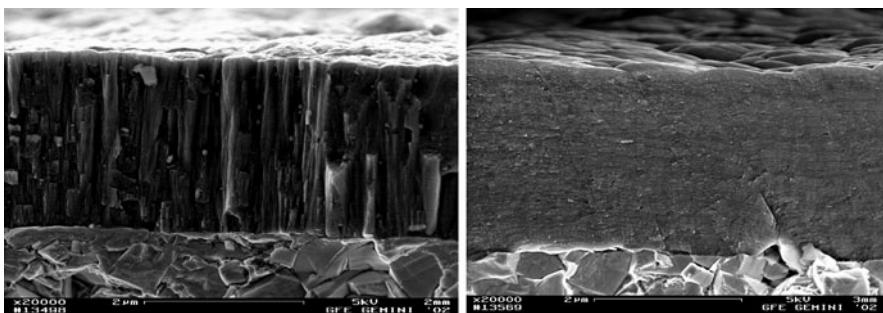
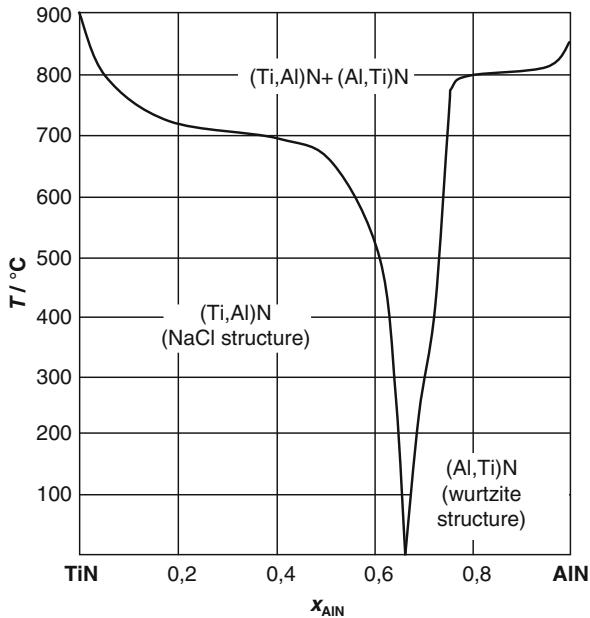


Fig. 4.38 Fracture pattern of (Ti,Al)N-layers on cemented carbide substrates, deposited with conventional DC-technique (left) and using bipolar-pulse-technique (right) (Source: CemeCon)

Fig. 4.39 Metastable phase diagram of the system TiN-AlN, acc. to Cremer [Crem98]



Ion Plating

With respect to process control, ion plating is defined as a variant of PVD coating, in which the condensing layer is altered by means of an additional electric field applied to the substrate. This method variant is independent of the type of process and physical principle involved in creating the metal vapour. The prerequisite is a sufficiently high number of ionised steam and process gas atoms.

In ion plating, the substrate is coated with a negative voltage or “bias voltage”. The metal vapour is ionized by means of electrodes arranged in a gas space and electromagnetic fields. Part of the ionised particles are accelerated towards the substrate (Fig. 4.40). Ion plating thus involves surrounding the substrate with gas ions. By bombarding the growing layer with ion, coating properties such as density and residual stress are modified and adhesion to the substrate is improved. The resultant coatings are very adhesive and tight. All PVD methods in which a bias voltage is used to modify the coating are termed “ion plating” [Bobz05].

In order to produce carbides, nitrides or oxides for coating purposes, here too a reactive gas is fed into the chamber. This method is called reactive ion plating [Möhwi96, Haef87]. Ion plating can also be utilized for purifying and activating the component surface prior to coating (also called etching) or for activation during coating [Rass96].

A variety of ion plating plant designs have emerged as practicable coating variants. For tool coating, arc ion plating (AIP) and ion plating with low-voltage arc discharge are among the process variants that have become the most significant in tool coating.

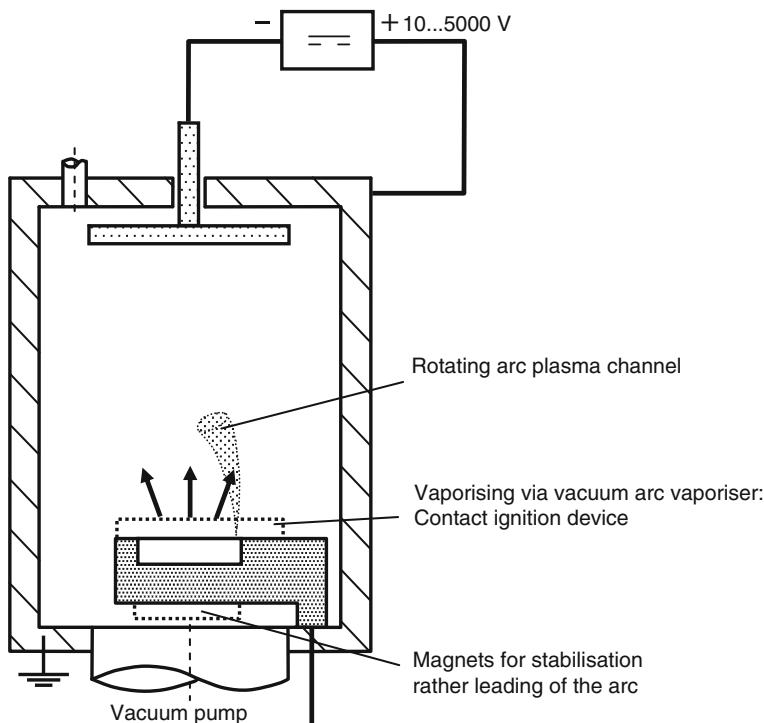


Fig. 4.40 PVD-Process–Ion plating, acc. to MÖHWALD [Möhw96]

Arc Ion Plating (AIP)

In the AIP process, the target material is vaporized and ionised by means of an arc in high vacuum. This method is characterized by the high degree of ionization of up to 90% and can thus be used to produce high quality coatings with dense structures and good adhesion.

The pre-voltage applied to the substrate accelerates the ionised particles, which impact the substrate at high speed. Their energy is partially converted into heat, which can result in high coating temperatures (approx. 550°C) in the substrates.

Magnetron Sputter Ion Plating (MSIP)

MSIP guarantees a stable plasma and high atomization rates [Bobz00]. Coatings deposited with MSIP exhibit very good mechanical and optical properties. Nearly all materials are depositable with this method. Since the ionisation degree of the atomized particles is under 5%, the substrates are heated to a much smaller degree than in the AIP process, so that plastics and materials with low annealing temperatures can also be successfully coated [Bobz00].

Low-Voltage Arc Coating

Ion plating with low voltage arc discharge is characterized by a low acceleration voltage, with which an electron jet is created from a low-voltage arc discharge. This jet is directed towards the target (anode) in a water-cooled crucible. The vapour made of the coating material arising from the electron bombardment is about 50% ionised and, since the substrate is applied with a voltage of -200 V, directed towards the coating target. This method is ideal for depositing nitride or carbide layers by introducing the corresponding reactive vapours [Haef87].

No droplets are formed when ion plating with low-voltage arc discharge is used for coating. This method thus makes it possible to deposit very smooth layers ($R_z = 0.1 \mu\text{m}$) and is thus suitable for coating mirror-polished surfaces without the need for post-polishing.

The Use of PVD-Coated Cemented Carbides at Low Cutting Speeds

Broaching soft and hardened steels offers an impressive example for the cutting edge durability of PVD-coated cemented carbides at low cutting speeds [Köni92a]. Characteristic of the broaching process are the generally low applicable cutting speeds, usually $1\text{--}25 \text{ m/min}$ depending on the broaching machine, cutting tool material and the target material.

Built up edge formation is characteristic of machining relatively soft or tough steels at low cutting speeds. These lead to dimensional and shape inaccuracies as well as to reduced workpiece surface quality. One measure for reducing built up edge formation among others is the use of broaching oils.

By using PVD-TiN-coated cemented carbides in place of uncoated ones, built up edge formation is almost completely suppressed due to lowered adhesion between the flowing chip and the cutting tool material in the entire cutting speed range under investigation of $v_c = 10\text{--}63 \text{ m/min}$ (Fig. 4.41). Furthermore, the use of coolants – a constant ecological problem whose maintenance and disposal are ever costlier – is superfluous.

4.4.2 Specific Properties of Hard Material Coatings

The most commonly utilized wear-protection coatings in metal machining commercially are TiN, TiC, Ti(C,N), Ti(Al,N), CrN and Al_2O_3 . Besides these coating systems, those that are characterized by a low friction coefficient such as DLC (diamond-like carbon) and MoS_2 -coatings are successfully used for wear-protection. Due to the unsurpassed hardness of diamond, CVD diamond coatings are also being increasingly used [Lemm03].

Different coating systems can be deposited as monolayers (e.g. TiN, Ti(C,N), Ti(Al,N), (Ti,HF,Cr)N) or, to improve their wear-protective effect, combined as multilayers with single layer thicknesses in the micrometers (e.g. TiN-Ti(C,N)- Al_2O_3 -TiN, (Ti,Al)N-WC/C).

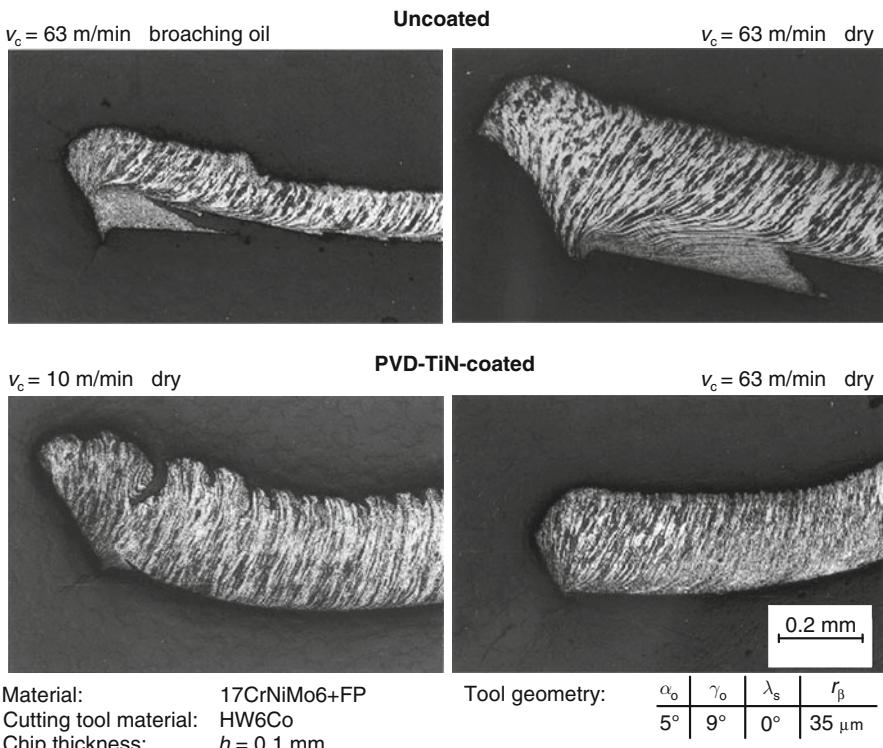


Fig. 4.41 Coating prevents formation of built up edges

The primary goals of multilayer technology are not only to improve the toughness, hardness and wear resistance of the coating but also to increase the total thickness of the coating. The transitions that are formed between layers of different materials place limits on shifting, which leads to a large increase in strength, especially in the case of coatings with very diverse Young's modulus. For cracks that are growing in the coating, transitions between the individual layers signify barriers at which they are forced into energy-dissipating detours, slowing or even stopping further growth (Fig. 4.42). The multilayer structure leads, in comparison to an equally thick monolayer coating, to lower residual stresses. Multilayer coatings also make it possible to integrate layers with different properties within one coating system, thus combining the functionalities of diverse coating materials. Such specific coating functions include, for example, high thermal stability, high hot hardness or the low friction resistance of intermediate or cover layers [Krie01, Lugs03]. Larger layer thicknesses mean a larger coating volume that has to be worn. From this fact results the generally longer standing times as well.

One further development in multilayer technology is the deposition of nanolayers or coatings with superlattice structures. These are coating systems consisting of a large number of extraordinarily thin layers, the thickness of which amounts to

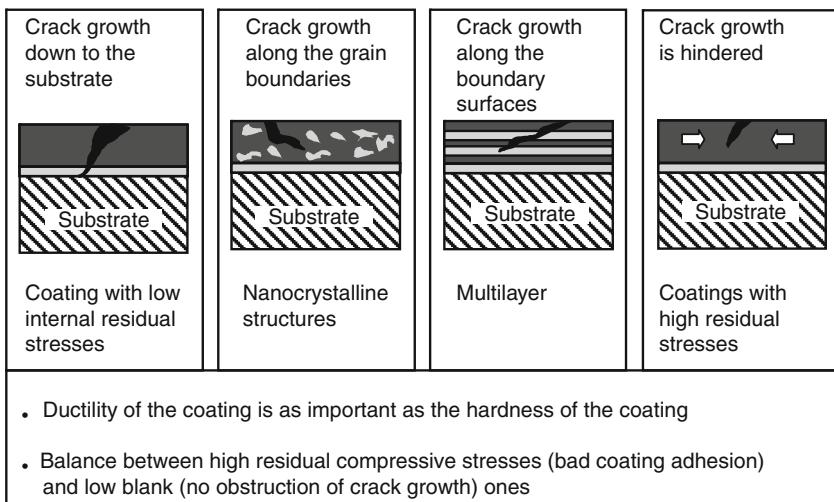


Fig. 4.42 Schematic diagram of the crack growth in hard coatings depending on structure and properties (Source: Oerlikon Balzers)

only a few nanometres (Fig. 4.43). In contrast to conventional, usually columnar layers, coating materials deposited as nanolayers are characterized by their significantly higher hardness. As has already been explained in the context of the PA-CVD coating process, this is based on the fact that the hardness of a material increases significantly below a certain layer thickness. This increase in hardness is explained by the phenomenon that diminishing layer thicknesses have altered crystal lattices and thus Young's modulus as well. For example, AlN has a hexagonal lattice structure at layer thicknesses > 10 nm and a cubic structure when the layer thickness is < 10 nm [Csel03a].

Another way to improve a coating's properties is to produce so-called nanocomposites. These are noncrystalline isotropic multiphase systems, in which two mutually insoluble phases (e.g. Al, Ti, Si) are deposited during the coating process on the tool surface. Examples of this are the embedment of nanocrystalline cubic TiN in an AlN matrix, of nanocrystalline TiAlN/AlCrN in an amorphous Si_3N_4 matrix [Csel04] or of nanocrystalline (Ti,Al)N into a matrix composed of (Al,Ti)N. The nanocomposites can be deposited as monolayer or multilayer coating systems. The boundary layers/grain boundaries in the nanolayer/nanocomposite layers are energy-dissipating barriers to cracks. Cracking and the speed of crack growth are thereby reduced (Fig. 4.42). Both nanolayer and nanocomposite layers are thus characterized not only by extremely high hot hardness and high-temperature wear resistance but also by favorable toughness attributes.

With respect to the structure of hard material coatings, we differentiate between monophase layers (e.g. TiN), multiphase layers (e.g. TiN + Ti₂N) and graded layers, i.e. layers the chemical composition of which exhibits a gradient (e.g. (Ti,Al)N layers with an increasing Al content) [VDI3824].

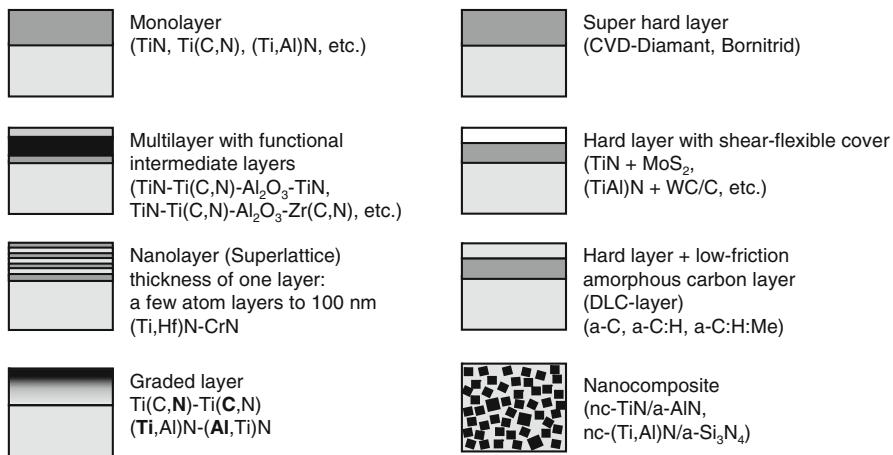


Fig. 4.43 Schematic diagram of the structure of current CVD and PVD layer systems

The properties of hard material coatings are determined by their chemical composition and structure, which depends on the conditions of their deposition. Figures 4.25 and 4.44 show some characteristic values for layer properties for some exemplary hard material coating systems produced by PVD and CVD processes. By changing the deposition conditions, such properties as chemical composition, morphology, structure, texture, residual stresses and thus microhardness, the thermal expansion coefficient and oxidation resistance can be altered within certain limits. The data presented can thus only be considered as guides for possible areas of application. Besides the hard material coatings here presented and explained in more detail below, there is a number of other hard material layers used for tool coating that are not more fully described here.

4.4.2.1 Titanium Carbide Coatings (TiC)

Coating tools with hard materials began with the deposition of TiC. Due to the high hardness realizable with titanium carbide (3100–3400 HV0.05), it provides more effective protection against abrasive wear than TiN. On the other hand, its tendency to diffusion is a bit higher than that of TiN and Al₂O₃ due to the relatively small enthalpy of formation. TiC's resistance to flank face wear is thus higher than that of TiN, while its resistance to crater wear is lower. Its oxidation wear is also the lowest of the layers shown in Fig. 4.44. TiC coatings are chiefly applied using the CVD process. For the most part, TiC is used as a monolayer when abrasion is predominant or in multilayer coatings together with TiN, Ti(C,N) or Al₂O₃ [VDI 3824].

4.4.2.2 Titanium Nitride (TiN)

Since 1980, tools were already being coated with titanium nitride. TiN is the most frequently employed hard material for coating cutting tools today. The coating material is an intercalation compound composed of titanium and nitrogen.

	TiN	TiCN	TiC	TiAlN	CrN	Al ₂ O ₃
Production process	PVD/CVD	PVD/CVD	CVD	PVD	PVD	CVD/PVD
Coating thickness/μm	1 to 5	1 to 5	1 to 5	1 to 5	1 to 10	1 to 5
Microhardness/HV 0,05¹⁾	2300	3000	3100	3000	1900	2100 HV 0.1
Oxidation temperature/$^{\circ}\text{C}$²⁾	> 450	> 350	> 350	> 700	> 600	_ ³⁾
Thermal barrier effect⁴⁾	+++	++	+	++++	+	+++++
Resistance to abrasion	++	+++	+++	+++	++	++
Resistance to wear due to adhesion (against steel)	++	++	+	++	++	+++
Resistance to wear due to diffusion (to steel)	++	+	+	+++	++	+++
Protection of basis material against corrosion⁵⁾	+	+	+	+	++	+

1) For the microhardness, mean values are indicated. They are obtained from reported measured values resulting from different compositions, coating thicknesses and internal stresses.

2) The oxidation temperature is the temperature at which oxidation of the coating material begins, considerably affecting the characteristics of the coating.

3) Al₂O₃ is already an oxide.

4) Being poor conductors of heat, the coatings act as a thermal barrier to the heat produced during metal-cutting so that most of the heat can be removed through the chip.

5) As the aforementioned hard coatings themselves do not corrode, they protect the basis material from corrosion. (Leaks in hard coatings may lead to the development of local galvanic elements and to pitting corrosion).

Fig. 4.44 Quantities Characterising the coating and the coating performance, acc. to [VDI 3824/1](#)

Due to the high interaction between the metal and nitrogen atoms, this compound is highly stable. With an enthalpy of formation that is almost twice as high as that of TiC, TiN is thermodynamically more stable, and thus more diffusion resistant and less inclined to adhesion. Therefore, TiN's resistance to crater wear is higher than that of TiC. TiN coatings are characterized by a high level of toughness.

TiN can be deposited with both the CVD and PVD process. The characteristic colour of TiN is goldish yellow. Silver-hued TiN coatings consist of a phase composite of Ti₂N and TiN, whereby the Ti₂N phase is dominant. These coatings are harder but also more brittle than pure TiN layers. They are most commonly used when abrasive wear predominates [[VDI 3824](#)].

4.4.2.3 Titanium Carbonnitride Coatings (Ti(C,N))

TiC and TiN can be mixed at any ratio. The properties of titanium carbonnitride are adjustable by varying the C/N ratio. With increasing carbon content, the colour of the Ti(C,N) layers change from coppery to violet, bluish-grey to grey [VDI 3824].

Titanium carbonnitrides are used industrially both as hard materials in cemented carbides and as wear-resistant thin-films. Often Ti(C,N) coatings are multilayered, i.e. deposited with increasing carbon content in the direction of the coating surface. By integrating carbon atoms in place of nitrogen atoms into the titanium nitride crystal lattice, a considerable increase in hardness can be realized, which is positive for wear resistance but also increases brittleness [Sato78, Schi74, Berg90]. To compensate for this increase in brittleness, Ti(C,N) coatings are deposited as multilayers, so that residual stresses between the individual coating layers can be reduced.

Ti(C,N) coatings are suited for machining steels with high tensile strength and thus for higher cutting temperatures.

4.4.2.4 Titanium Aluminium Nitride Coatings ((Ti,Al)N)

The (Ti,Al)N coating system was developed in order to improve the oxidation resistance, hot hardness and wear-protection properties above the levels of previously used coatings [Quin87, Knot87]. In comparison to TiN and Ti(C,N) coatings, (Ti,Al)N coatings have the highest oxidation resistance with a comparably high level of hardness. Since Ti(Al,N) is a metastable coating system, it can only be deposited with the PVD, PA-CVD or MT-CVD process.

(Ti,Al)N coatings are a further development of TiN, whereby titanium is substituted by aluminium by 20–60 at.-%. Depending on the composition, these layers range from brown (lower Al-content) to black-violet (higher Al-content) [VDI 3824].

Due to their high level of oxidation resistance and hot hardness, the preferred areas of application are dry machining, hard machining and HSC machining. The excellent wear resistance of (Ti,Al)N coatings is explained by the fact that, as opposed to stable composite phases, metastable layers decompose into stable boundary phases or form stable oxides in an oxidizing atmosphere as long as the energy required to convert into an equilibrium condition is added (e.g. as heat), as it is the case in dry machining or high-speed machining. In this context, the high oxidation resistance of (Ti,Al)N is derived from the fact that a thin aluminium oxide layer is formed on the coating surface which is constantly renewed during the cutting process, thus decelerating the progress of wear. With increasing Al-content, the oxidation resistance of the (Ti,Al)N coating is increased. Despite its much further improved oxidation resistance compared to TiN and Ti(C,N) coatings, this coating also fails beyond approx. 800°C. TiAlN coatings are deposited as monolayer, multilayer or gradient layers [Lemm03, Leye04, VDI 3824].

(Ti,Al)N is one of the most commonly used high-performance coating systems. Important measures being taken to further augment the cutting edge durability of (Ti,Al)N coatings include the development of nanolayers and increasing the content of aluminium. The primary goal of these developments is to improve their hot

hardness, wear resistance and oxidation resistance. In order to differentiate them from conventional (Ti,Al)N coatings, those with over 50% aluminium content are designated as (Al,Ti)N coatings.

4.4.2.5 Al-Cr-N Coatings

Besides increasing the aluminium content another approach to further improving the oxidation resistance and high-temperature properties of (Ti,Al)N coating systems is adding small amounts of oxide formers such as Cr, Y or Si. One example for this direction in coating development is the Al-Cr-N coating system. In comparison to the conventional (Ti,Al)N coating, the (Al,Cr)N coating system has proved to have a higher resistance to abrasive wear as well as higher hot hardness and oxidation resistance [Gey04]. Its areas of application include the dry, HSC and hard machining of steel and non-ferrous metals by turning, drilling and milling [Gey04, Denk04].

4.4.2.6 Aluminium Oxide Coatings (Al_2O_3)

The exceptional resistance of Al_2O_3 against both abrasion and diffusion wear and its simultaneous insensitivity to oxidative wear, known from its use in cutting ceramics, has made Al_2O_3 an obvious candidate as a hard material for coating [Sche88].

Because it is very brittle, Al_2O_3 is generally not used in monolayer form, but only in combination with other hard materials in multilayer coatings. The electrically nonconductive Al_2O_3 layers could thusfar not be fabricated with the PVD, but only with the CVD process for technical reasons. The deposition of the thermodynamically stable high-temperature modification of aluminium oxide – $\alpha\text{-Al}_2\text{O}_3$ – by means of CVD process has been established industrially for over two decades. Due to its excellent wear protection and high performance potential, attempts are being made to produce Al_2O_3 layers with the PVD process as well. A first step in this direction was the deposition of amorphous aluminium oxide ($\alpha\text{-Al}_2\text{O}_3$) with the help of RF sputtering technology. Because of the low coating rates and consequently long process durations, this technique is proven to be uneconomical for coating cutting tools. With the help of the pulsed magnetron sputtering process, it has become possible to coat cutting tools with crystalline $\gamma\text{-Al}_2\text{O}_3$ layers at substrate temperatures of 500–600°C. $\gamma\text{-Al}_2\text{O}_3$ can be applied either as a monolayer or multilayer coating in combination with carbidic or nitridic intermediate layers [Hauz05].

4.4.2.7 Amorphous Carbon Coatings

The concept “amorphous carbon coatings” includes a number of coatings and coating systems. They are often referred to as “diamond-like carbon” or “DLC” coatings. These are carbon-based, highly cross-linked amorphous layers with different amounts of sp^2 (graphite bond) and sp^3 bonds (diamond bonds) as well as various amounts of embedded hydrogen. In the case of Me-C:H layers, finely distributed metal carbides are also embedded. DLC coatings are manufactured by means of plasma-activated PVD and CVD techniques and deposited at coating temperatures ranging from room temperature to about 300°C. Their typical layer

	MeC:H	a-C:H	a-C	Diamond
Production process	PVD	CVD	PVD	CVD
Coating thickness in μm	1 to 10	1 to 5	1 to 3	3 to 10
Microhardness in HV 0,05¹⁾	800 to 1800	1500 to 3500	3000 to 7000	10000
Internal stresses in GPa	0.1 to 1.5	1 to 3	2 to 6	–
Graphitisation temperature in $^{\circ}\text{C}$²⁾	350	400	450	> 600
Resistance to abrasion	+	+++	++++	++++
Resistance to wear due to adhesion (against steel)	+++	+++	+++	(++) with good cooling
Protection of basis material against corrosion³⁾	+	+++	+++	+++

1) For the microhardness, mean values are indicated. They are obtained from reported measured values resulting from different compositions, coating thicknesses and internal stresses.

2) The graphitising temperature is the temperature at which amorphous carbon begins to convert from a three-dimensional lattice into graphite, considerably affecting the characteristics of the coating.

3) As the aforementioned hard coatings themselves do not essentially corrode (except for MeC), they protect the basis material from corrosion. (Leaks in hard coatings may lead to the development of local galvanic elements and to pitting corrosion.)

Fig. 4.45 Quantities Characterising the coating and the coating performance, acc. to [VDI 3824/1](#)

thicknesses are 1–5 μm . Figure 4.45 shows some characteristic values for the layer properties of the most important carbon-based hard material layers producible with PVD and CVD available today [[VDI 3824](#), [VDI 2840](#)].

In order to improve their adhesive strength, carbon layers are as a rule deposited on a hard material layer (e.g. CrN) as a top layer. Graded intermediate layers made of CrCN have proved favourable for this, in which case hardness and the Young's modulus can be adjusted to the mechanical properties of the carbon top layer by successive exchange of nitrogen and carbon. This procedure makes it possible to obtain a largely continuous transition from the hard material layer to the carbon layer in order to optimize the bond between both coating systems [Leye04].

Amorphous carbon layers are characterized by extremely low friction coefficients in the case of solid body friction, high wear resistance and low tendencies to adhesion. In cutting technology, these coatings are used above all on tools for dry machining NE metals (e.g. aluminium or magnesium) and in fine machining with great success. Further areas of application are in machining graphite and fibre-reinforced plastics.

The property profile of amorphous carbon layers can be purposefully influenced by process control and by integrating additional chemical elements. This potential has led to a number of diverse types of carbon coatings in the last several years. In the following, the essential characteristics and properties of some selected amorphous carbon layers will be described in close detail. The classification, designation and characterization of the coatings are derived from VDI guideline 2840.

According to VDI guideline 2840, amorphous carbon coatings can be subdivided into two groups, hydrogen-free and hydrogenous coatings. Since all carbon layers contain a certain amount of hydrogen even without adding hydrogen gas (e.g. from residual gasses), a limit of about 3 at.-% hydrogen is seen as the transition from hydrogen-free to hydrogenous carbon coatings. With increasing amounts of hydrogen, the amount of interconnection between the carbon atoms decreases, leading to softer layers [VDI 3824, VDI 2840].

As wear-protection coatings on cutting tools, hydrogen-free carbon layers are used above all in the form of the amorphous carbon layer a-C and the tetraedic amorphous carbon layer ta-C. The atoms are arranged randomly in amorphous solid bodies. The only bonds are those between a few isolated atoms. If the sp^2 bond is predominant in the amorphous carbon layer, it is softer, while it is harder if the amount of sp^3 is higher. Depending on the deposition energy, the carbon atoms in the layer are predominantly arranged in one of the two hybrid states. At low deposition energies, the amount of sp^2 bonds is higher, making the layers softer. These have the abbreviation "a-C". With high deposition energies, sp^3 hybridizations with a tetraedic arrangement predominate. The layers exhibit a high hardness level, and the compressive residual stresses are increased. Tetraedic amorphous carbon coatings have the designation "ta-C" [VDI 3824, VDI 2840].

Of all amorphous carbon coatings containing hydrogen, it is above all the non-modified layers that besides carbon contain only hydrogen, designated with "a-C:H" as well as modified carbon layers that are of any importance. The latter contain further elements in addition to hydrogen and are classified in accordance with VDI 2840 into the group of carbon layers containing metal and the group of hydrogenous amorphous carbon layers modified with non-metals [VDI 3824, VDI 2840].

The a-C:H coating system represent the origin of coating with amorphous hydrocarbon layers. In addition to carbon, it contains 10–30% hydrogen. The latter is derived from hydrocarbon gases like acetylene, which are employed in the fabrication process.

If, during vacuum deposition, metals are additionally integrated, layers containing metal are produced. As opposed to a-C:H layers, these are electrically conductive and can be manufactured with a technically simpler and cost-efficient DC method. Of all amorphous carbon layers, these therefore have the broadest and deepest range of application. Amorphous carbon layers containing metal are also generally known as "a-C:Me" or "a-C:H:Me" (the abbreviation Me-C:H is also common), where "Me" stands for "metal". Instead of the abbreviation "Me", the integrated metals can also be named specifically, e.g. tungsten (a-C:W, a-C:H:W) oder titanium (a-C:Ti, a-C:H:Ti). The added metals form finely distributed carbides

with the carbon in the matrix. By integrating metal carbides, layer adhesion can be improved and the tribological properties of the coating can be affected [VDI 3824, VDI 2840].

Hydrogenous amorphous carbon layers modified with non-metals contain non-metallic elements such as silicon (Si), oxygen (O), nitrogen (N), fluorine (F) or boron (B), which partially also form carbides. By incorporating different non-metals, further improvements become possible. Silicon for example helps to increase temperature resistance. Moreover, different elements can be built into the layer at the same time. In this way, special layer properties like surface energy (adhesion tendency, wettability) can be altered.

4.4.2.8 Self-Lubricating Coatings

Self-lubricating coatings include coating systems which include graphite or molybdenum disulfide (MoS_2). On cutting tools, graphite and MoS_2 coatings are applied as top layers on a supporting hard material layer, where they act as a solid lubricant. Graphite is deposited in combination with WC as a WC/C multilayer coating system consisting of several layers of WC and graphite.

Graphite and MoS_2 have similar crystal structures. It is characteristic of graphite that its C atoms are arranged in a plane resulting in a very high level of internal bond strength due to the small inter-atomic distance. There is a relatively large distance between the atom layers however, so the layers of atoms can be easily shifted against each other under the influence of external forces. MoS_2 also has a lamellar structure. There is a very strong chemical bond between the Mo and S atoms forming each lamella. The bond between two neighbouring lamellae is made between the sulphur atoms. Due to the intermolecular van der Waals bonds between the sulphur atoms of two lamellae, these can be shifted towards one another very easily. These slip planes are responsible for the soft, lubricating effect of graphite and molybdenum sulphide.

Both coating systems are used primarily for machining materials that tend to adhere strongly with the tool, for example in the case of dry machining aluminium wrought alloys. In the case of drilling, these layers decrease the friction between the chip and the tool because of their “lubricating” effect and thus help to improve chip removal. In the case of tapping, in which case the machining results are determined above all by crushing, friction and adhesion processes, these coating systems can bring about significant improvements in performance even without hard material intermediate layers.

Because of their properties, both coating systems are gradually worn out during the cutting process (loss lubrication). Even if the layer appears optically to be worn, its effect is still partially demonstrable. This is explained by the fact that the coating material still exists in microscopically small recesses on the tool surface where it can still be effective. After a complete loss of the soft top layer, the further wear properties of the tools is then determined by the properties of the supporting hard material system.

4.4.2.9 Diamond Coatings (Crystalline Carbon Coatings)

Coating tools with polycrystalline diamond has been available only since 1990 and is as such the most recent technique in diamond tool production. Diamond coatings are applied to tools made of cemented carbide or ceramics in a low-pressure diamond synthesis (a CVD process). The diamond layer consists of pure diamonds and does not have a binder phase. Industrial deposition temperatures are around 600–1000°C. This coating technology makes it possible to manufacture even tools with complex geometries, such as curved surfaces, economically out of diamond [Leye95]. Application examples for this are drilling tools, end milling cutters and indexable inserts with chip form geometries. These tools are used in the machining of plastics, cemented carbide and ceramic green bodies, non-ferrous metals as well as in the machining of abrasive materials. Diamond-coated tools can not be used to machine steel and other ferrous materials since the diamond coating is easily worn due to the solubility of carbon in iron.

We differentiate between thin diamond layers ($s_d = 1\text{--}40 \mu\text{m}$) and thick diamond layers ($s_d = 0.3\text{--}2 \text{ mm}$). Thin diamond layers are used to coat components (tools, for example) directly. On the other hand, thick diamond layers are generally deposited on an auxiliary substrate and then detached from it again. Diamond plates produced in such manner are then used as free-standing diamond (e.g. as radiation windows) or mounted on supports, usually by means of vacuum soldering in order to produce tools or other components. Since this can no longer be referred to explicitly as a layer, we often no longer refer to it as a CVD diamond layer but as a CVD diamond [VDI 2840].

Several processes have been established for fabricating diamond coatings. The most important include the hot filament process (HF-CVD), the microwave process and the plasma jet process (DC arcjet). What all of these processes have in common is the decomposition of carbonic gases (e.g. methane, acetylene or carbon monoxide) followed by a deposition of carbon on the substrate surface. The formation of unwanted graphite must be avoided in the process. Atomic hydrogen plays a key role in the germ formation and growth of the diamond layers. In the case of the hot filament method, the best-known process used for depositing CVD thin diamond layers, the atomic hydrogen required for the process is produced by electrically heated filaments made of refractory metals (Ta, W or Re). Gas temperatures of over 2500°C are necessary to initiate the required chemical reactions [Lemm04, VDI 2840].

The rim zone is of particular importance when coating cemented carbides. Cobalt, which forms the binder phase of the cemented carbide, reduces the germ formation of diamond on the substrate surface, accelerates graphite formation, and affects crystal growth and layer adhesion. Cemented carbide substrates must therefore be subjected to pretreatment prior to coating, in which case we differentiate between physical and chemical pretreatment and the introduction of an intermediate layer. The common process presently is a multi-stage pretreatment consisting of irradiation, chemical etching, purifying and seeding. The substrate is first purified and homogenized with microbeam treatment. In the subsequent chemical etching

process, the cobalt binder phase is removed from the rim zone of the cemented carbide substrate. The depth to which the binder phase is removed depends on the back diffusion of cobalt during the coating process. Due to the high substrate temperature of 900°C and long processing times, cobalt diffuses through the etched rim zone to the substrate surface during the diamond coating process. If the etching depth is too small, the cobalt that has diffused to the surface reacts with the diamond layer and leads to the separation of diamonds on the boundary surface. This lowers the adhesive strength of the diamond coating. If the etching depth is too large, the carbide bonds in the binder phase are weakened, also lowering layer adhesion. The rough surface caused by etching serves mechanically to clamp the diamond layer to the cemented carbide substrate. Due to the extreme differences between the coating and substrate material (Young's modulus and hardness) this mechanical clamping is of special importance with respect to the adhesive strength of the diamond layer. After etching, the substrates are again purified and seeded. In order that the diamond layer will develop, there must be diamond germs on the substrate surface that serve as starting crystals and from which the layer grows. Seeding can be performed, for example, by applying small diamond crystals with a diamond suspension [Gram04, VDI 2840, Lemm04].

Standard diamond layers have a microcrystalline structure (Fig. 4.46). The crystallites grow from the germs with varying speed depending on their crystal orientation. Slowly growing crystallites are covered by the faster-growing ones. With increasing layer thickness, the crystallites forming the surface become larger and larger, resulting in the typical appearance of a microcrystalline CVD diamond layer with sharp-edged crystal surfaces [VDI 2840].

If growth conditions are set such that new germs are constantly formed and large-scale growth of individual crystallites becomes impossible, crystallite size remains in the range of nanometres within the entire layer (Figs. 4.46 and 4.47). This is how “nanocrystalline” CVD layers are formed, which are much smoother compared to crystalline layers [VDI 2840].

Both layer types can also be combined in multilayer structures, uniting the positive attributes of microcrystalline and nanocrystalline diamond layers [Uhlm05] (Figs. 4.46 and 4.47). This increases the fracture toughness of the entire system and contains the development of cracks, such that they can no longer easily reach the

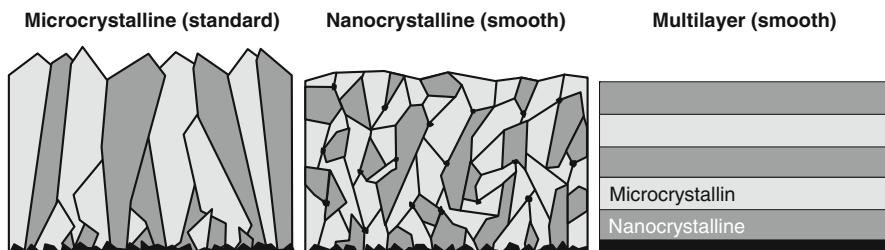


Fig. 4.46 Growth model of diamond coatings, acc. to VDI 2840 and LEMMER [Lemm04]

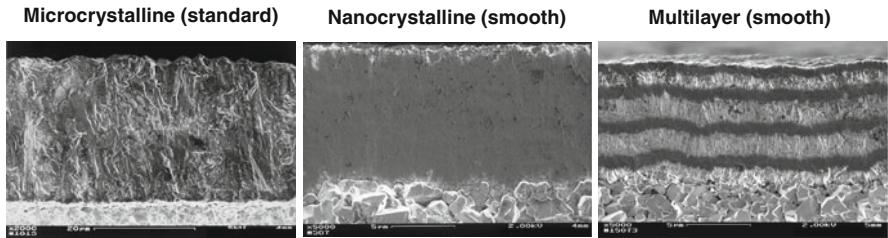


Fig. 4.47 SEM photographs of CVD diamond coatings on a cemented carbide substrate [VDI 2840, Lemm04]

interface between the first diamond layer and the substrate, potentially causing the entire surface to chip [Lemm04].

4.4.3 Substrate Pretreatment

Both CVD and particularly PVD coating processes place high demands on the surface conditions of the components that are to be coated. Even the smallest impurities on the surface of the substrate – be it residual cutting fluid or grease, water deposits, grinding and polishing agents, hardening salt residues or oxide deposits – can degas during the coating process, negatively influence that process and lead to coating adhesion problems. The pretreatment of the substrate therefore has a key role in creating high and reproducible coating quality [Lugs03, VDI3824].

In substrate pretreatment a distinction is drawn between mechanical pretreatment (grinding, deburring, microbeam and polishing) and physicochemical pretreatment (coating-removal, ultrasound-supported purification). In order to optimize layer adhesion, the parts are subjected to a multi-staged ultrasound purification process before charging. In the coating plant, the parts are then first heated in a vacuum, whereby further impurities can evaporate. Then the tools are etched by means of argon ion bombardment in order to produce a purely metallic surface. Coating follows directly thereafter. If the substrates still need to be stored temporarily before charging, this takes place in a drying closet in order to protect the substrate surfaces from water deposits (from air moisture), rust film formation and skin contact [Lugs03, VDI3824, Gey03].

4.5 Ceramic Cutting Tool Materials

Ceramic materials include all non-metallic, inorganic solid materials. Essentially these are chemical compounds of metals with non-metallic elements of group III A to group VII A of the periodic system of elements. A distinction is made between oxidic and non-oxidic ceramics. The largest group of ceramics are the oxides. Non-oxidic ceramics are carbides, borides, nitrides and silicones. Some authors differentiate again between metallic hard materials (compounds between C, B, N or

S and Ti, Zr, Nb, Ta, W among others) and non-metallic hard materials such as diamond, SiC, Si_3N_4 , B_4C and BN. Characteristic properties of ceramic materials are compressive strength, high chemical resistance and high melting temperatures, which can be derived from the strong covalent and ionic bonds of the atoms [Salm83a, Horn06].

The following will describe more closely those ceramic materials that are used as cutting tool materials in cutting technology. Following the usage common in practice, we will differentiate between cutting ceramics (oxidic and non-oxidic ceramics) and superhard non-metallic cutting tool materials (diamond and boron nitride).

4.5.1 Cutting Ceramics

Ceramics have applications in many areas of machining with geometrically defined cutting part geometries. Mentionable among these are the high-speed machining of cast iron materials, hard machining or nickel-based alloy machining. The use of ceramics in the mass production of brake disks, flywheels and like workpieces has become a matter of course. In the field of automotive technology, silicon nitride ceramics have been able to capture an important market share in the high-speed cutting of cast iron workpieces [Schn99]. Hard machining is a primary application area of mixed ceramics. “Whisker-strengthened” ceramics are often used in the area of engine construction for machining engine components made of nickel-based alloys. Whiskers are needle-shaped monocrystals with a small degree of misorientation in the lattice. Correspondingly, they have high mechanical strength (R_m up to 7000 N/mm^2). Their length is about $20\text{--}30 \mu\text{m}$, their diameter between 0.1 and $1 \mu\text{m}$.

The increased interest in ceramics in production is based not only on the excellent wear properties of ceramic cutting tool materials but most importantly on their toughness properties, which have been clearly increased in recent times. The brittle fracture attributes characteristic of ceramic materials, the wide variation of their strength properties and the resulting stochastically appearing tool fractures are as before the main causes limiting the use of these cutting tool materials in machining technology (in contrast to cemented carbides for example).

New developments and improvements in ceramics have been concentrated therefore on further augmentation of the toughness and dependability of these cutting tool materials in production.

In the case of pure oxidic ceramics, research is focused on improving these properties by increasing their zirconium oxide content, improving the distribution of these phases and by making the microstructure more consistent. In the case of mixed ceramics, finer-grain hard materials are used and titanium carbide is being partially substituted with titanium carbon nitride. Further measures include fibre and/or whisker reinforcement, reduction of the amount of glass phase at the grain boundaries of non-oxidic ceramics as well as the formation of special microstructures [Kola86, Clau77, Schn99].

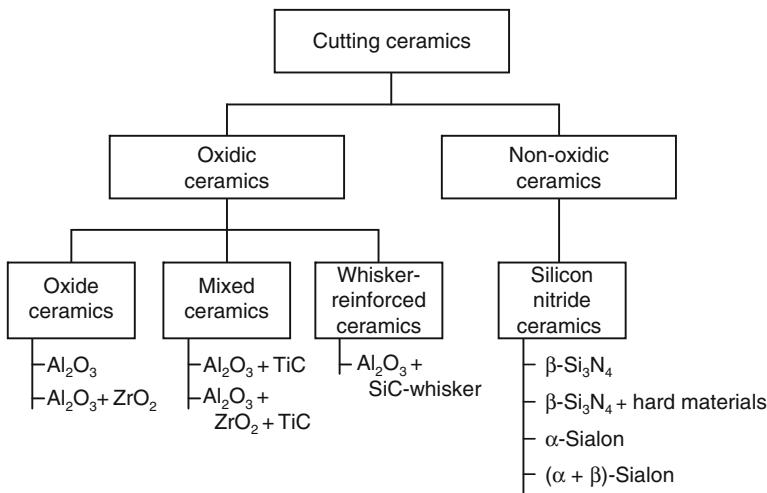


Fig. 4.48 Classification of cutting ceramics

Ceramic cutting tool materials can be subdivided into oxidic and non-oxidic cutting ceramics (Fig. 4.48).

Oxidic cutting ceramics include all cutting tool materials based on aluminium oxide (Al_2O_3). A distinction is made between oxide ceramics that only contain oxides (e.g. ZrO_2) in addition to Al_2O_3 , mixed ceramics, which in addition to Al_2O_3 also contains metallic hard materials (TiC/TiCN) and whisker-reinforced ceramics, in which SiC whiskers are integrated in the Al_2O_3 matrix.

Figure 4.49 shows images of fractures of different cutting ceramics taken with a scanning electron microscope. It is typical of oxidic cutting ceramics that they do not have a visible binder phase such as cemented carbides (cobalt) have. One can recognize the globular grains of oxide ceramics, the extremely fine grains of mixed and whisker-reinforced ceramics as well as the SiC whiskers, which are only a few μm large in the detail magnification. Characteristic of non-oxidic ceramics based on Si_3N_4 is the needle-shaped formation of the crystals. In order to improve wear resistance, Si_3N_4 cutting ceramics can be coated with Al_2O_3 or with multilayer coating systems made of Al_2O_3 and TiN [Schn99].

Figure 4.50 shows some physical properties of commercially available ceramic cutting tool materials. Depending on their chemical composition, these ceramic types can have considerably diverse properties.

4.5.1.1 Cutting Ceramics Based on Al_2O_3

Oxide Ceramics

White oxide ceramics are the traditional type of cutting ceramics. Ceramics based Al_2O_3 were already introduced as cutting tool materials at the end of the 1930s [Kola86]. Inserts were made for a long time out of pure aluminium oxide. Due

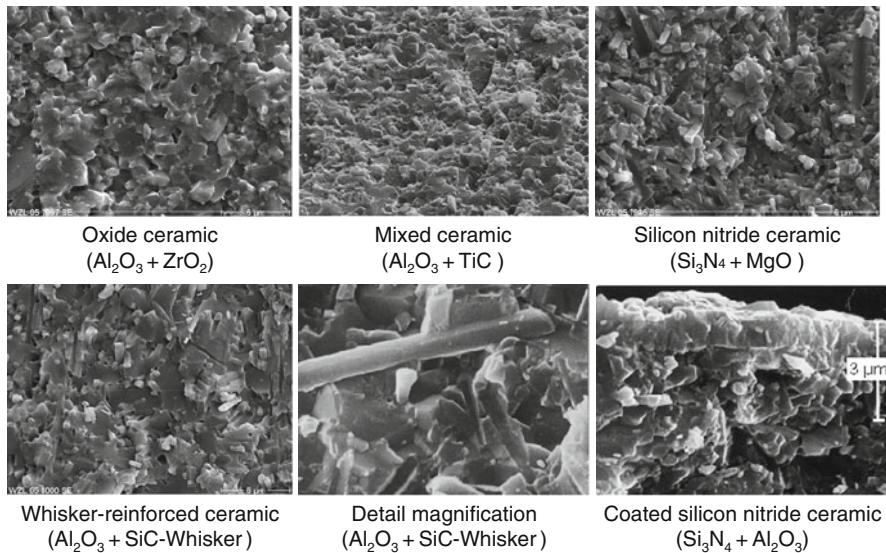


Fig. 4.49 Fractures of cutting ceramics

Properties			Oxide ceramic Al_2O_3	Whisker-reinf. Oxide ceramic Al_2O_3 +15% ZrO_2 +20%SiC-Whisker	Mixed ceramic Al_2O_3	Silicon nitride ceramic Si_3N_4
Density	ρ	g/cm ³	4.0	4.2	3.7	4.1
Vicker hardness	-	-	1730	1750	1900	1730
Bending strength	σ_{bB}	N/mm ²	700	800	900	650
Compressive strength	σ_{dB}	N/mm ²	5000	4700	-	4800
Young's modulus	E	$10^3 \cdot \text{N/mm}^2$	380	410	390	390
Fracture toughness	K_{IC}	N·m ^{1/2} /mm ²	4.5	5.1	8.0	4.2
Thermal conductivity	λ	W/m·K	16.4	15	32	14.7
Thermal expansion	α	$10^{-6} \cdot \text{K}^{-1}$	8	8	-	8
Melting point	T	°C	-	-	-	-

Fig. 4.50 Physical and mechanical properties of different cutting ceramics and their main components

to their brittleness and vulnerability to fracture, such inserts are no longer used in machining. The pure ceramics used today are dispersion material containing Al_2O_3 as well as about 3–15% of finely distributed zirconium dioxide to improve toughness.

The toughness-enhancing effect of dispersed ZrO_2 particles in an Al_2O_3 matrix is based on the phase transformation of zirconium dioxide. ZrO_2 , which exists in the form of a tetragonal lattice modification in the sintering temperature range (1400–1600°C), transforms during cooling into its monoclinic low-temperature modification. The temperatures in which this transformation takes place depend on the size of the particles. The smaller the ZrO_2 particles are, the lower the transformation temperature. Since the transformation from the tetragonal to the monoclinic modification is associated with a volume expansion, various specific mechanisms of action can assert themselves depending on the size of the particles. The common effect of all these mechanisms is that they ultimately absorb fracture energy. The speed of crack development is reduced by microcracking, crack branching, the stress-induced transformation of small ZrO_2 particles as well as crack diversion. The result of this is that critical cracks only develop at a higher level of energy, which corresponds to an increase in fracture resistance and an improvement of ductility [Clau77, Clau84, Zieg86].

Mixed Ceramics

Mixed ceramics (black ceramics) are dispersion materials based on Al_2O_3 that contain between 5 and 40% of non-oxidic components in the form of TiC or TiCN. The hard materials in the matrix form finely distributed phases, which limit the growth of aluminium oxide grains. Correspondingly, these ceramics have a very fine-grained structure, improved toughness properties and a high level of edge strength and wear resistance. Compared to pure ceramics, they are harder and have more favourable thermoshock properties due to their high level of thermal conductivity (Fig. 4.50). The toughness of these ceramics can be further improved by adding ZrO_2 .

The development of mixed ceramics is progressing towards finer and finer-grained cutting tool materials with extremely homogeneous textural structures. As opposed to conventional cutting ceramics based on aluminium oxide and titanium carbon nitride with an average grain size of $< 2 \mu\text{m}$, the more advanced mixed ceramics have a submicron structure with grain sizes of $< 1 \mu\text{m}$. The finer-grained structure increases hardness and bending strength and consequentially the mechanical and thermal loadability, wear resistance and edge strength of ceramic inserts. Fine-grained mixed ceramics are used in hard-fine machining, e.g. for machining hardened rolling bearing steels, for hard-fine turning case-hardened automotive components such as drive wheels, crown wheels, gearwheels or sliding sleeves with a Rockwell hardness of 54–62 HRC but also for planing and fine planing cast iron at very high speeds. In the case of hard turning, submicron mixed ceramics are competing with PCBN cutting tool materials in many areas of application as more economical alternatives due to their good cost-benefit ratio [Krel97, Schn99].

Whisker-Reinforced Cutting Ceramics

Whisker-reinforced cutting ceramics are cutting tool materials based on Al_2O_3 with about 20–40% silicon carbide whiskers. The goal of whisker-reinforcement is to enhance the toughness properties of ceramic cutting tool materials. The increase in toughness gained by incorporating whiskers in oxidic ceramics is remarkable. Compared to mixed ceramics, whisker-reinforced types have up to 60% higher fracture toughness. The whiskers bring about a more consistent distribution of mechanical loads in the cutting tool material as well as a more rapid transport of heat from the thermally highly loaded cutting areas due to their improved heat conductivity. This results in improved fatigue limits and thermoshock resistance, so that whisker-reinforced cutting ceramics can also be used in wet cut processes.

Fabrication of Oxidic Cutting Ceramics

The starting materials of cutting ceramics, including pressing and sintering auxiliaries, are metered according to an exact formula, homogenized in vibration mills and converted to a compressible powder with a high bulk material stability in a spray-dryer.

Oxide ceramics, as well as mixed ceramics with small amounts of hard materials, are compressed at room temperature to inserts (cold pressing) and then sintered. The sintering temperature is about 1600°C. Mixed ceramics with a high percentage of hard material (> 10%) as well as whisker-reinforced ceramics have to be hot-pressed, i.e. pressing and sintering take place in one cycle. Due to the special equipment required for this (e.g. graphite moulds instead of cemented carbide ones), hot pressing is more cost-intensive than normal sintering. Furthermore, the number of manufacturable tool geometries is much more limited. Hot isostatic pressing can be used to subsequently compact sintered ceramic parts, thus reducing the porosity of the cutting tool material. One method variant is hot isostatic compacting of pre-sintered closed-pore mouldings in an installation.

After sintering/hot pressing, the contact face, and in the case of precision inserts, the lateral surface and cutting edge chamfer are ground with diamond grinding discs [Kola86, Grew85].

Properties of Oxide Ceramic Cutting Tool Materials

Because of their high hot hardness and chemical resistance, ceramic cutting tool materials based on Al_2O_3 are characterized by excellent wear properties. However, the highly favourable wear properties of oxide ceramic cutting tool materials is accompanied by their sensitivity to tensile, bending, impact and thermal shock stresses. As mentioned at the beginning of this chapter, current further development of oxide ceramic cutting tool materials has for this reason been concentrated on improving toughness and thermoshock properties without diminishing wear resistance.

The excellent wear resistance of oxidic cutting tool materials, particularly at high temperatures, is based among other things on their superior hardness and compressive strength compared to other cutting tool materials at high temperatures. For example, oxide ceramic cutting tool materials at 1000°C are still harder than high speed steel at room temperature (Fig. 4.51).

While the compressive strength of Al_2O_3 at room temperature roughly corresponds to that of cemented carbide, it is at 1100°C still as great as that of steel at room temperature, while neither steel nor cemented carbide are resilient to pressure at 1100°C. The high compressive strength of ceramics has been a primary reason for their being privileged as cutting tool materials. But because their banding fracture strength is relatively low, the use of oxide ceramics as the cutting tool material requires that the process kinematics be selected such that the cutting forces are effective as much as possible only in the form of compressive stresses [Kola86, Kola86a, Zieg86].

A further property which makes the use of ceramic cutting tool materials advantageous at high cutting speeds is Al_2O_3 's low level of creeping. In the case of cemented carbides, high temperature strength is limited by the property-determining cobalt phase at temperatures of 800–900°C. At higher temperatures, creeping processes are introduced which are smaller by several orders of magnitude in the case of Al_2O_3 .

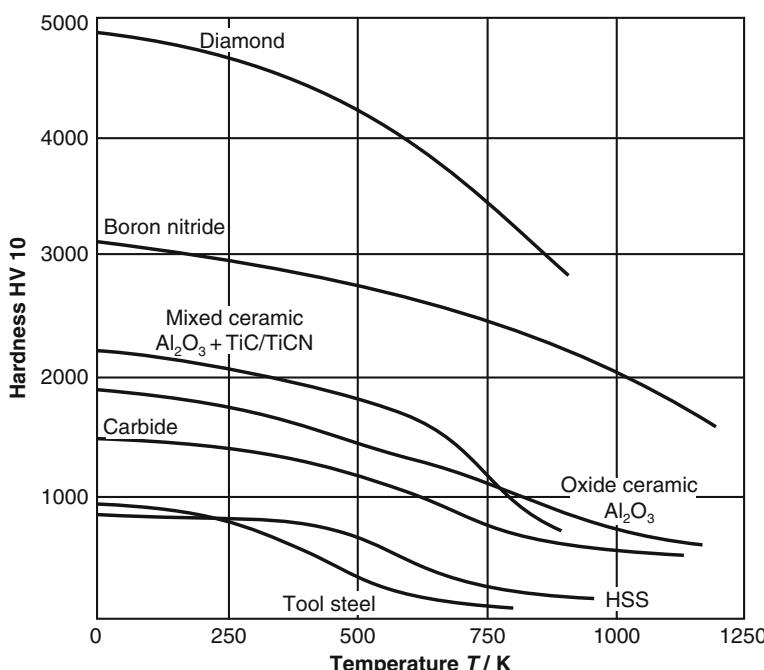


Fig. 4.51 Hot hardness of cutting tool materials (Source: CeramTec)

The high wear resistance of oxitic cutting tool materials can also be attributed to the good chemical resistance of Al_2O_3 . Al_2O_3 is resistant to oxidation at the cutting temperature used in practice and has only a small amount of affinity to metallic materials.

The most crucial disadvantage of ceramic cutting tool materials is their brittleness, i.e. lacking capacity to reduce stress peaks by means of plastic deformation. The cause of this is the low number of sliding systems in the crystal structures of ceramic materials. This results in high sensitivity to tensile stresses and low resistance to mechanical and thermal shock compared with metals.

In the case of a sudden mechanical overload due to impulse-like stress, the cutting tool material is destroyed by brittle fracture. Because of insufficient ductility, cracks begin to form when the inner material cohesion is exceeded by external stress. If the crack reaches a critical size, unstable crack growth appears leading to fracture of the ceramic tool. The stress intensity factor K_I is a parameter for judging the stress condition at the peak of a crack. Failure of the cutting tool material due to fracture occurs if the stress intensity factor reaches a critical value, crack toughness (fracture toughness) K_{IC} . In comparison to other materials, the values of K_{IC} are very low for ceramics (Fig. 4.50).

Another important disadvantage of oxide ceramics are their relatively low resistance to temperature change. In the case of temperature changes of more than 200°C , pure Al_2O_3 is destroyed. This can only be improved by alloying a component of superior resistance to temperature change.

Due to their sensitivity to thermoshock, coolants should not be used in conjunction with oxide ceramics when roughing and planing. Coolants should only be used to temper the workpiece, e.g. due to narrow workpiece tolerances.

Low bending fracture strength and relatively high sensitivity to impact and temperature change stresses require a slanted entering and leaving cutting path, which is associated with delayed load of the cutting edge of the tool (Fig. 4.52). Due to their low edge strength, chamfered edges should stabilize the cutting edge.

Areas of Application of Oxide Ceramic Cutting Tool Materials

It must be stressed that, due to their extreme brittleness, a careful adjustment of cutting parameters to tool geometry is the prerequisite of a successful use of cutting ceramics. There is no room for error, since ceramics can be destroyed immediately in the form of fracture should the marginal conditions be unfavourable. Cutting ceramics are therefore almost exclusively used in mass production. This is justified by the reduction of processing times and the high cost of selecting and adjusting the cutting tool materials. The main area of application of oxide ceramic cutting tool materials is rough and finish turning of grey cast iron, case-hardened steels and heat-treated steels. However, pure oxide ceramics are being increasingly replaced by silicon nitride ceramics, especially in cast iron processing (Fig. 4.53). The increased popularity of dry machining however is opening up new areas of application to oxide ceramics in the case of turning extrusion parts with a small machining allowance.

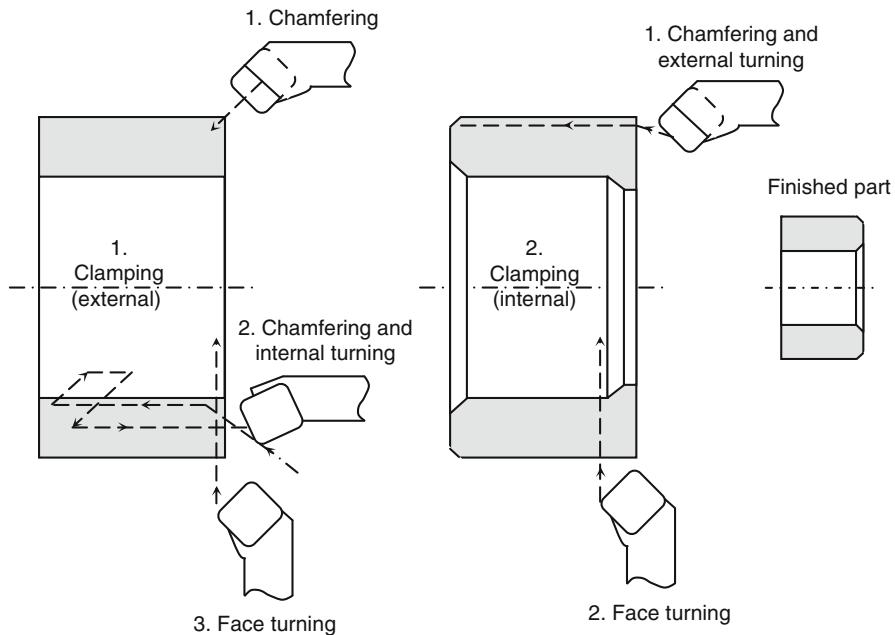


Fig. 4.52 Entering and exit tool path when turning with cutting ceramics (Source: Ford)

Fine-grained mixed ceramics are used in hard-fine machining (e.g. for machining hardened rolling bearing steels) for fine turning case-hardened automotive components such as drive wheels, crown wheels, gearwheels or sliding sleeves with a Rockwell hardness of 54–62 HRC. They are also used for finish planing and fine planing cast iron at very high speeds. In the case of hard turning, submicron mixed ceramics are competing with PCBN cutting tool materials in many areas of application as more economical alternatives due to their good cost-benefit ratio [Krel97, Schn99].

Both types of cutting ceramics still however remain restricted with respect to the carbon content of the machined steels. For example, in the case of steels with a C content of under 0.35% adhesion occurs on the tool as well as chemical reactions which increase wear and generally make the process uneconomical.

Whisker-reinforced oxide ceramics have been used with great success for turning high temperature nickel-based alloys (Fig. 4.54). In contrast to otherwise used cutting tool materials (cemented carbide or HSS), it is possible to increase the cutting speed by a factor of 10 or more. In the case of rough and finish milling Inconel 718 with milling head face cutters, cutting speeds of 800–1000 m/min can be realized with cutting ceramics based on silicon nitride or whisker-reinforced aluminium oxide. As opposed to machining with cemented carbide tools, this represents an increase in performance by factors of 25 or more [Momp93, Gers02, Krie02, Uhlm04].

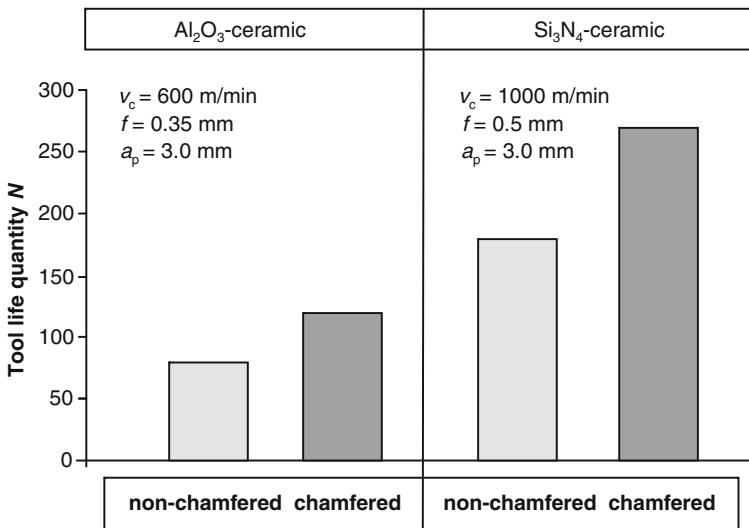


Fig. 4.53 Influence of cutting ceramics and entering path on the tool life quantity in brake disc manufacturing (Source: CeramTec)

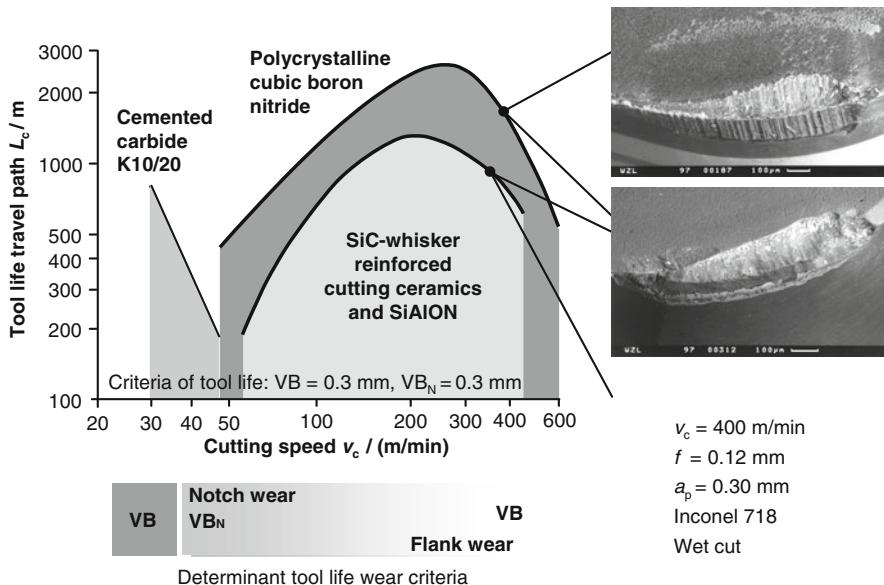


Fig. 4.54 Achievable tool life travel paths and applicable cutting speeds in finish turning Inconel 718 with uncoated cemented carbide, whisker-reinforced cutting ceramic, silicon nitride ceramic and PCBN

Chemical reactions and built-up edge formation when machining light metal alloys make Al_2O_3 cutting ceramics unsuitable for machining Al, Mg and Ti alloys.

Cutting ceramics composed of thin ceramic tubules represent a new generation of ceramic cutting tool materials. In the case of these ceramics, first introduced at EMO 2005, the tool body is formed of small thin tubes consisting of a tough ceramic material and filled with a hard, wear-resistant ceramic material. The diameter and thickness of the tubes, as well as their orientation and filling, are variable. The tubes are arranged in a plane. The tool is formed with a large number of planes built on top of each other, where the tubes forming the individual planes can be parallel from plane to plane or can be arranged alternately at an angle of $45^\circ/90^\circ$ to each other. In this way, the properties of the cutting tool material, such as toughness and wear resistance, can be modified and adjusted to the stresses of various machining tasks. Initial machining experiments in turning high-temperature nickel-based alloys as well as in machining hardened steel materials demonstrated the promising wear properties and performance of this new cutting ceramic [Haid05].

4.5.1.2 Non-oxidic Cutting Ceramics

Among the non-oxidic ceramics (carbides, nitrides, borides, silicides...), materials based on Si_3N_4 have been especially successful as cutting tool materials for machining purposes. Si_3N_4 cutting ceramics, in contrast to oxidic cutting ceramics, are characterized by increased toughness and improved thermoshock resistance (Fig. 4.50). Moreover, they have a higher hot hardness and high temperature strength. In machining processes involving grey cast iron, they make it possible to use the highest cutting values while maintaining the highest tool life spans and low failure rates. The high safety level of these cutting tool materials has been an especially important factor in the acceptance of Si_3N_4 cutting tool materials by manufacturers.

The significantly higher fracture resistance compared to oxide and mixed ceramics is based on the needle-like shape of the hexagonal $\beta\text{-Si}_3\text{N}_4$ crystals as opposed to the globular Al_2O_3 grains. The non-directional growth of the needle-shaped crystals leads to a microstructure made up of mechanically interlinked components which provides the cutting tool materials with excellent strength properties. The low thermal expansion of Si_3N_4 ceramics compared with oxide and mixed ceramics is responsible among other things for its favourable thermoshock properties. However, in order to press silicon nitride ceramics completely, sintering auxiliaries (Y_2O_3 , MgO , Al_2O_3) are required that form a glass phase and fill in the gaps between the crystals. This glass/binder phase has a negative influence on the high-temperature properties of silicon nitride ceramics.

Besides sintering auxiliaries, Si_3N_4 cutting ceramics can also contain other additives that affect their crystal structure or texture and thus their properties as well. Corresponding to their chemical composition and crystallographic structure, the silicon nitride ceramics available today can be subdivided into three groups:

- I: $\beta\text{-Si}_3\text{N}_4 + \text{binder phase } (\text{Y}_2\text{O}_3, \text{MgO}, \text{Al}_2\text{O}_3)$
- II a: α sialon ($\alpha\text{-Si}_3\text{N}_4 + \text{Al}_2\text{O}_3 + \text{AlN} + \text{binder phase } (\text{Y}_2\text{O}_3)$)

- II b: ($\alpha + \beta$) sialons ($\alpha \text{Si}_3\text{N}_4 + \text{Al}_2\text{O}_3 + \text{AlN}$) + additive (e.g. ytterbium) + binder phase (Y_2O_3)
- III: β silicon nitride + hard materials (e.g. TiN, ZrO_2 , SiC-whiskers) + binder phase

The cutting tool materials from Groups I and II are fabricated by means of hot pressing, sintering, hot isostatic pressing or a combination of these methods. Most of the Si_3N_4 cutting tool materials available on the market today belong to Group I.

The cutting tool materials from Group II are usually designated as sialons. Silicon nitride can receive up to 60% aluminium oxide in solid solution. Some nitrogen atoms are replaced in the process by oxygen atoms and silicon atoms by aluminium atoms. While the α sialon mixed crystals have a globular form, β sialon mixed crystals are stem-shaped. Including special additives makes it possible to stabilize the β sialons, thereby creating a cutting tool material the structure of which consists of both α and β mixed crystals. The amount of additives determines the amount of β sialons in the cutting tool material's microstructure. In comparison to the cutting tool materials from Group I, sialons are harder, more chemically resistant and have an increased resistance to oxidation. The manufacturing method in this case is sintering, which can sometimes be followed by hot isostatic pressing.

Group III can include Si_3N_4 cutting tool materials, the properties of which can be specifically altered by adding hard materials such as titanium nitride, titanium carbide, zirconium oxide or Sic whiskers.

Areas of Application of Non-oxidic Cutting Ceramics

The classic field of Si_3N_4 cutting tool materials is grey cast iron machining. In this case, usually the tougher silicon nitride ceramics of Group I are preferred, especially in automated manufacture. Due to the high fracture toughness of these cutting tool materials, large feeds, high cutting speeds and thus large volume removal rates can be realized when machining cast iron materials in smooth and interrupted cut [Köni86, Köni87, Schn99]. For example, when turning automotive brake discs, using Si_3N_4 ceramics improves tool life quantity considerably compared with oxide ceramics (Fig. 4.53). Improved resistance to fracture makes it possible, as shown in Fig. 4.52, to omit workpiece chamfering as long as one forgoes maximal utilization of the cutting tool material.

The main area of application of sialons is in turning nickel-based alloys under finishing or average roughing conditions (Fig. 4.54). They are therefore in direct competition with whisker-reinforced oxide ceramics. Like the latter, sialons can be used for machining nickel-based alloys at much higher cutting speeds than cemented carbide tools. They are also good for high-speed milling of nickel-based alloys with tools mounted with indexable inserts at cutting speeds of 800–1000 m/min.

The wear resistance of silicon nitride ceramics is slightly lower than that of oxide ceramics. Silicon nitride cutting tool materials exhibit a strong affinity to iron and oxygen under machining conditions. They wear very quickly in the case of steel machining, so the use of these materials groups currently remains uneconomical.

Coating silicon nitride ceramics opens up further prospects for their use. The classic coating materials are TiN, TiC, TiCN and Al₂O₃, which are applied in different combinations and layer thicknesses. A multilayer coating system consisting of aluminium oxide and titanium nitride remains the most popular. The service life of Si₃N₄ can be considerably increased by applying a wear and diffusion inhibiting coating, especially when machining cast iron with globular graphite (e.g. GJV40) [Schn99]. Coated silicon nitride ceramics are also suitable therefore for cases that had hitherto been reserved for cemented carbides. As opposed to cemented carbides, they allow for higher cutting speeds and thus for shorter machining times and in many cases also for dry machining [Schn99].

4.5.2 Superhard Non-metallic Cutting Tool Materials

In machining technology, cutting tool materials based on diamond and boron nitride are designated as superhard and non-metallic. According to the definition, both are ceramic cutting tool materials, where diamond is assigned to monatomic, boron nitride to non-oxidic ceramics.

4.5.2.1 Diamond as a Cutting Tool Material

Elemental carbon appears in the two crystal modifications graphite and diamond. Diamond solidifies in the cubic crystalline lattice system, in which the C atoms are bonded covalently and tetrahedrally. The extremely high bond and lattice energy is the reason why diamond is the hardest of all known materials. Diamond is equally superior to all other known hard materials with respect to thermal conductivity. Figure 4.55 shows the hardness and thermal conductivity of some materials that tend to be used as abrasives as well as different carbides, which are components of cemented carbide tools and permits a comparison with diamond.

Classification of Diamond Cutting Tool Materials

To classify diamond cutting tool materials, we differentiate between natural and synthetic diamond, which both can appear in monocrystalline or polycrystalline form. According to DIN ISO 513, the identification letters of polycrystalline diamond are DP while those of monocrystalline diamond are DM (Fig. 4.3). Following the usage common in practice and in the literature, we will also refer to polycrystalline diamond cutting tool materials with the abbreviation PCD.

(a) *Natural Diamond*

Natural diamond is only of importance for the purposes of cutting with geometrically defined cutting edges in its monocrystalline form. Although diamond also exists in nature in polycrystalline form (ballas, carbonado), these types of diamond are of lesser interest since the synthetic fabrication of polycrystalline diamond is more advantageous both economically as well as technologically.

One significant property of monocrystalline diamond is the anisotropy (directional dependence) of mechanical properties such as hardness, strength

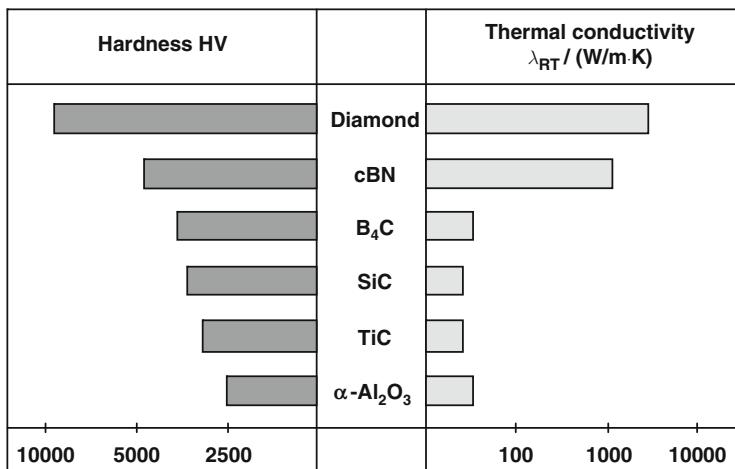


Fig. 4.55 Hardness and thermal conductivity of diamonds in comparison to other hard materials

or elastic modulus. While this directional dependence is compensated in the case of polycrystalline materials by the completely random distribution of the individual crystals, anisotropy remains in monocrystalline materials due to the aligned orientation of the crystal lattice.

This is also the cause of the cleavability of monocrystalline diamonds in four select cleavage directions. From this we can see that the position of the lattice directions must be known both for grinding monocrystalline diamond as well as for using it as a tool. While grinding must always take place in the direction of lowest hardness, monocrystalline diamond tools have to be oriented in the tool holder such that the cutting force points in the direction of a hardness maximum.

(b) Synthetic Diamond

The fabrication of synthetic diamonds takes place using a catalyst solution in a pressure and temperature range in which diamond forms a stable phase. By a careful selection of pressure and temperature, the crystal growth rate and dimensions of the crystals can be controlled within a range of a few micrometers to several millimetres and specific physical properties like purity or porosity can be influenced.

Diamond synthesis results in monocrystalline diamond particles that are utilized in machining with geometrically defined cutting edges as synthetic monocrystalline diamond tool bodies or for further processing into polycrystalline cutting parts.

Fabrication and Designs of Diamond Tool Bodies

(a) Monocrystalline Diamond Tools

Monocrystalline diamond tool bodies are manufactured with both natural diamond and synthetic monocrystals [Heus96]. The most common cutting edge shapes of

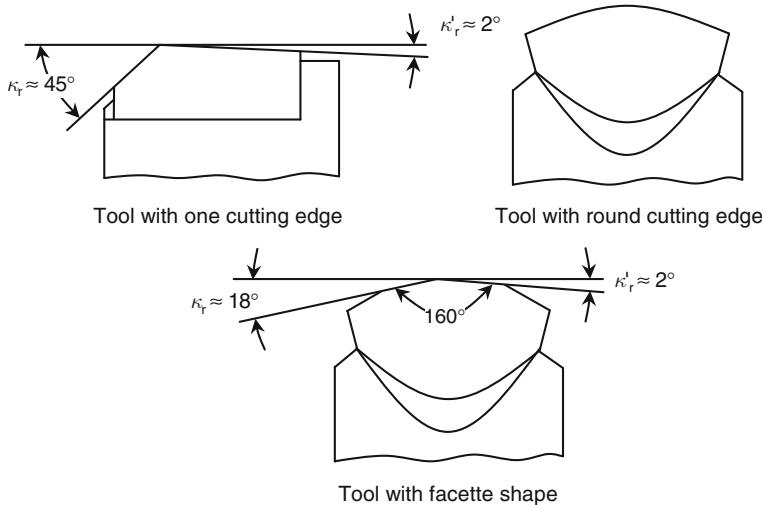


Fig. 4.56 Shapes of monocrystalline diamond tools

monocrystalline diamond tools are shown in Fig. 4.56 [Wein69]. The tool with one cutting edge is used both for drilling as well as for external turning, whereby in the latter case a corner radius is ground onto it in order to improve the surface profile. By selecting a very small minor cutting edge lead angle $\kappa'_r < 2^\circ$, the minor cutting edge can be designed as a broad-tool cutting edge which takes on the function of further polishing the surface.

Tool variants with round cutting edges have the advantage of having a very large useful cutting edge length, but due to the curvature of the cutting edge they have unfavourable chip formation conditions and relatively high passive forces. In the case of tool bodies with facette shapes, three to five cutting edges are ground on, whereby two neighbouring edges form an angle of about 160° . The tool is adjusted so that the minor cutting edge has a very small lead angle and acts as a broad polishing cutting edge. Surface quality can be affected considerably by varying κ'_r slightly. The grinded tips are brazed on the support material or fixed with special clamps.

(b) Polycrystalline Diamond Tools

Tool bodies made of a synthetic polycrystalline diamond layer were first introduced in 1973 and have replaced monocrystalline diamond tools and cemented carbides in some areas. The starting materials are synthetic diamond particles of a very small defined granulation (grain diameter $2\text{--}25\ \mu\text{m}$) in order to obtain a maximum amount of homogeneity and packing density.

The polycrystalline diamond coating is fabricated by means of a high pressure, high temperature process ($60\text{--}70\ \text{kbar}$, $1400\text{--}2000^\circ\text{C}$), in which synthetic diamonds are sintered together into polycrystalline bodies in the presence of a metallic catalyst. To this end, usually cobalt, but also silicon, tungsten or tungsten carbide are

used [Neis94]. During the sintering process, “diamond bridges” are formed between the diamond grains which give the polycrystalline bodies their high strength. As the diamond grains consolidate, gaps arise between neighbouring diamond crystals that are filled in by the catalyst. In the literature, this phase is also called the binder, which gives the polycrystalline diamond bodies the required toughness [Pret06]. One way to reduce the size of the gaps between the diamond grains – especially in the case of coarse-grained DP types – is using diamond particles with varying grain sizes (2–30 μm). The small diamond grains accumulate in the gaps formed by the large crystals, they increase the number of diamond bridges and with it the impact resistance of the composite material as well as the quality of the cutting edge. The polycrystalline coating fabricated in the high pressure/high temperature process, the thickness of which is about 0.5 mm, is either directly applied on a presintered cemented carbide base or over a thin intermediate layer consisting of a metal with a low elastic modulus bonded with the cemented carbide in order to balance stresses between the diamond layer and the cemented carbide base.

Due to its polycrystalline structure, the diamond coating is a statistically isotropic overall body in which the anisotropy of the individual monocrystalline diamond particles is balanced by the random distribution of the diamond grains. Polycrystalline diamond thus does not exhibit the hardness isotropy and cleavability of monocrystalline diamonds. On the other hand, it also does not reach the hardness level of a diamond monocrystal in its “hardest” direction, especially since hardness is also influenced by the degree of consolidation between the individual crystals and their bonding to the binder phase.

Depending on the grain size of the diamond granulation used as well as the catalyst used, the properties of the polycrystalline composite material can be specifically influenced. Coarse-grained DP cutting tool materials (grain size about 25 μm) are characterized by higher hardness and thermal resistance, but their cutting edges are rounder and more jagged than those of finer-grained types. As a result, wear resistance increases in the case of primarily abrasive stress during the machining process (Fig. 4.57). They are the most advantageous with respect to wear, cutting edge quality and cutting edge strength in the machining of aluminium composite materials and alloys, which are not critical with respect to the workpiece surface. DP types of average grain size (approx. 10 μm) are very common as multi-purpose variants. They are used for rotating tools with high requirements on cutting edge quality and service life. Fine-grained types (about 2 μm grain size) have established themselves especially in the automobile industry in cases demanding the best possible cutting edge quality [Bail01].

Changing the composition of the “binder phase” can also increase the limits of thermal stability, for example by using SiC. However, this considerably reduces the thermal conductivity of the cutting tool material compared with that of materials with the usual Co catalyst [Neis94].

Cutting edge shaping of the cutting part blank (i.e. of the cemented carbide base with applied diamond coating) is accomplished by means of spark-erosive cutting and grinding. The tool bodies are either soldered onto the tool bearer or clamped into standardized tool holders, since indexable inserts made of polycrystalline diamond

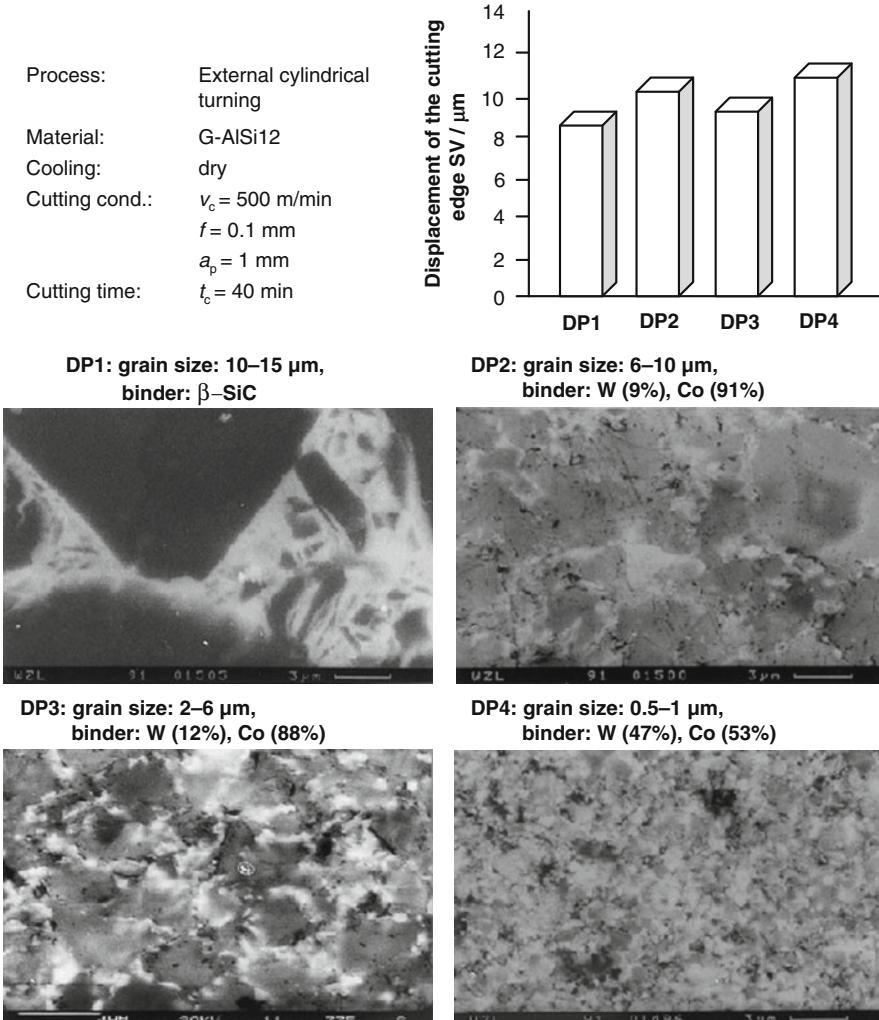


Fig. 4.57 Microstructure and wear of polycrystalline diamond cutting tool materials when turning AlSi-alloy G-AlSi12, acc. to NEISES [Neis94]

correspond in shape and dimensions with commercially available cemented carbide or ceramic indexable inserts.

Application Areas of Diamond Cutting Tool Materials

Cutting iron and steel materials with diamond tools is impossible due to the affinity of iron to carbon. Diamond turns into graphite and reacts with the iron in the contact zone between the tool and the workpiece due to the high temperatures arising there.

As a result, the cutting edge wears quickly in the case of both monocrystalline and polycrystalline diamond cutting tool materials.

Monocrystalline diamond tools are especially suited to cutting light, heavy and precious metals, hard and soft rubber as well as glass, plastics and stone. Their field of application is mainly in finishing, since large depths of cut and feeds are impossible due to limitations on cutting edge dimensions and relatively low flexural strength. The use of monocrystalline diamond cutting tool materials promises advantages when the demand for very high dimensional accuracy and surface quality is of foremost importance. For example, the use of nearly notch-free polished diamond cutting edges in ultra-precision cutting (turning, fly-cutting, planing) can result in surface finishes of between 3 and 6 nm. One classic area of application is machining hard and soft contact lenses. Further fields include the production of mirrors for lasers and other optical applications, of shaping tools for making impressions on blank CDs, of tools for making imprints on plastics or of tools for plastic injection moulding [Spen91, Ikaw91, Weck95, Heus96, Brin96, Kloc96, Take00].

Besides light, heavy and precious metals, the palette of materials machined with polycrystalline diamond tools comprises different plastics, coal, graphite and presintered cemented carbide. Their use is not restricted to finishing, but also includes roughing. It is in many cases possible to unite both pre-machining and finishing in one working cycle.

In the case of machining workpieces made of aluminium and other non-ferrous metals, PCD tools have acquired a secure place in modern cutting processes as high-performance tools [Schü01, Halw04, Kass04, Vogt04, Wick04, Fall05, Hedr05, Hedr05a, Kasp05, Brun99]. Polycrystalline diamond tools are of special importance in the machining of aluminium alloys containing large amounts of silicon (Fig. 4.58). Since these alloys have a hard/soft structure, the cutting edge cuts in an alternating fashion through the soft aluminium phase and through the hard silicon particles. Due to the strong abrasive effect of the silicon particles, tools made of cemented carbide are subject to a lot of wear. Moreover, when using cemented carbide tools, the adhesive tendency of aluminium with the cutting tool material has a negative effect on the machining process.

In comparison to cemented carbides, PCD cutting tool materials can be used with much higher cutting speeds when cutting aluminium alloys with high silicon content. They have extremely high tool lives of up to 80 times higher than those of cemented carbide tools and have excellent surface quality and precision. Their high level of manufacturing safety is another essential reason why tools made of polycrystalline diamond are preferred over those of cemented carbide for this machining task. This is especially the case in mass production on transfer lines, where short cycle times and high safety against unforeseen cutting edge fracture are a must. Further application examples include the milling of magnesium pressure cast alloys and the high-performance milling of aluminium integral parts in the airline industry with end milling cutters equipped with PCD [Wein69, Obel84, Köni82, Chry79, Spur84, Hoff88, Jäge89, Beck95, Wald92, Stie99, Zwah00].

Under certain conditions, it is possible to machine cast iron materials with PCD tools as well. One example of this is drilling finishing with reaming tools equipped

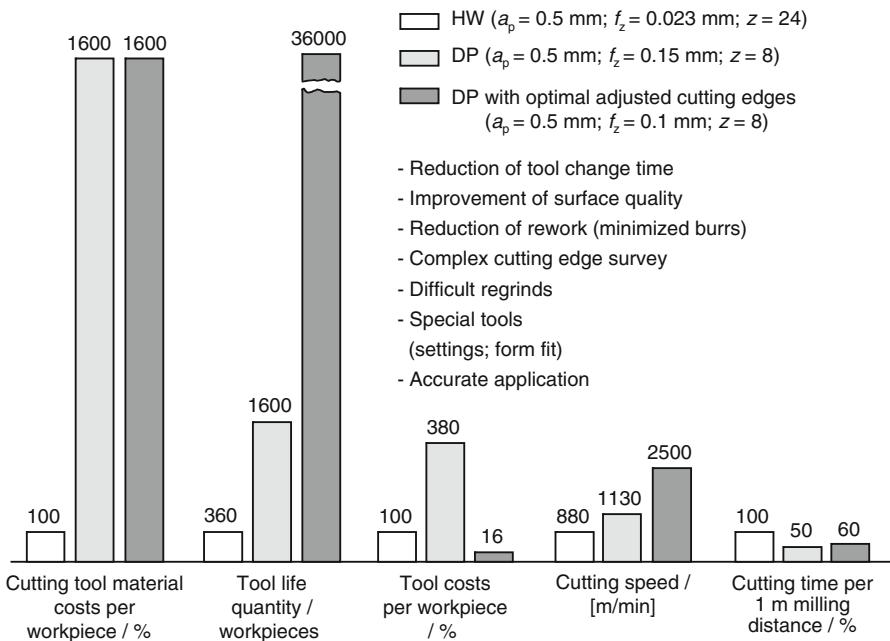


Fig. 4.58 Comparison of cemented carbide and polycrystalline diamond when milling the aluminium-based alloy GK-AlSi17Cu4Mg

with DP. In the case of very small depths of cut, adjusted feeds and cutting speeds as well as an intensive cooling of the cutting edge, it is possible to obtain long tool lives as well as high surface quality and component precision in cast iron machining [NN02].

(c) CVD Diamond Tools

CVD diamonds are diamond coatings deposited on a base body made of cemented carbide or ceramics in a CVD process. A distinction is drawn between thin and thick diamond coatings. As opposed to polycrystalline diamond, these diamond layers consist only of diamond crystals, containing no “binder phase”. The fabrication of tools coated with diamond is treated extensively in Sect. 4.4.2. Thin diamond coatings can be deposited on tools with complex geometries such as drills, end milling cutters and indexable inserts with chip form grooves. They are suited to machining graphite, copper, fibre-reinforced plastics and aluminium alloys high in silicon. In the field of superfinishing, thin diamond coatings have not been successful thus far. The reason for this is that sharp cutting edges are blunted by the relatively thick diamond coatings (about $20 \mu\text{m}$). Recent developments are making it possible to produce very sharp cutting edges on diamond-coated tools as well by means of a sharpening process [Hage04].

Thick CVD diamond coatings are excellently suited to cutting highly abrasive materials such as aluminium-based metal composite materials. Examples of this are

the machining of brake drums made of AlSi9Mg into which 20 vol% SiC particles are embedded or of cylinder crankcases made of AlSi9Cu3, which are reinforced with 15 vol% Si particles and 5 vol% Al_2O_3 short fibres [Suss01, Wein02, Uhlm00]. In the case of machining GFK/CFK materials as well, much longer tool lives can be expected compared with PCD cutting edges due to the lack of a binder phase [Feuc05].

4.5.2.2 Boron Nitride as a Cutting Tool Material

Boron nitride exists like carbon in a soft hexagonal modification, which crystallizes in the same lattice type, and in a hard cubic modification, which has an identical structure to the diamond lattice. There is also a third modification, which crystallizes in the wurzite structure. The wurzite lattice is a lattice type with a hexagonal symmetry but with a different atomic configuration than the graphite lattice. With respect to hardness, this form is somewhere between the other two modifications.

In contrast to silicon nitride, naturally existing hexagonal boron nitride is soft and not suitable as a cutting tool material for machining with defined cutting part geometries. Only after transforming the hexagonal into the cubic crystalline lattice with the help of a high-pressure/high temperature process does boron nitride exhibit those qualities that distinguish it as a cutting tool material (Fig. 4.55). Cubic boron nitride is, after diamond, the second hardest material. Hexagonal boron nitride is synthesized by reacting boron halogenides with ammonia. It has a density of 2.27 g/cm^3 and a melting point of 2730°C [Salm83]. Cubic boron nitride ($\rho = 3.45 \text{ g/cm}^3$) does not exist in nature. Its fabrication under the conditions of diamond synthesis was first successful in 1957. The transformation of hexagonal into cubic boron nitride is accomplished with pressures of 50–90 kbar and temperatures of 1800–2200 K under the catalytic influence of alkaline nitrides or alkaline-earth nitrides. Figure 4.59 shows the phase diagram of boron nitride system with the four phase areas molten, hexagonal boron nitride with graphite structure, hexagonal boron nitride with wurzite structure and cubic boron nitride with diamond structure.

Despite their identical lattice structures, there are essential differences between diamond and cBN. cBN has six cleavage planes – two more than diamond. This property is insignificant for the use of cBN in machining with defined cutting edges, since in that case only polycrystalline tools are used.

More important is the fact that boron nitride is not a chemical element like carbon, but a chemical compound. The boron nitride lattice contains boron and nitrogen atoms and can therefore not reach the same level of bonding force symmetry and hardness as diamond, the lattice of which consists exclusively of carbon atoms.

But with respect to its chemical resistance, especially against oxidation, cBN is far superior to diamond. It is stable from atmospheric pressure to about 2000°C , whereas diamond begins to graphitize already at about 900°C .

In the case of machining with geometrically defined cutting edges, cBN is mainly used as a polycrystalline cutting tool material (BN). Since the maximum crystal size in the fabrication of cBN is limited ($1\text{--}50 \mu\text{m}$), cBN grains are sintered with

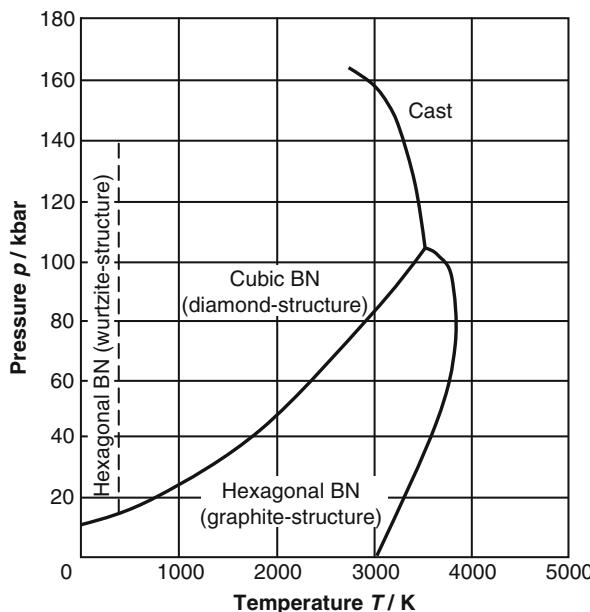


Fig. 4.59 Phase diagram of boron nitride

the help of a binder phase in a high pressure/high temperature process to an approx. 0.5 mm thick polycrystalline BN layer and simultaneously applied to a cemented carbide base. A distinction is made between cutting edge blanks, which are connected by soldering to the tool holder and only receive their final cutting edge shape by grinding, and indexable inserts, which are clamped in common clamping holders [Töns81, Töll81, Wern81, Nott82, Clau85, Töll82, Wern84, Wein87, Buzd87, Krat06].

Aluminium, titanium, titanium aluminium nitride, titanium carbide or cobalt, nickel and tungsten are used as binders. Aluminium and titanium react chemically with the cBN, forming AlN and AlB_2/TiN and TiB_2 . Cobalt forms complex borides with tungsten in the form of CoWB, W_2CoB_2 or $\text{W}_2\text{Co}_{21}\text{B}_6$. The coW borides arise from grain boundary reactions during the sintering process. This leads to an extremely stable bond between the cBN grains and the Co-W binder phase [Krat06].

The currently most common cutting tool materials based on boron nitride can be subdivided into two groups, one containing high amounts of cBN (BH) and the other containing low amounts (BL) (Tables 4.1 and 4.2). Types with high amounts of cBN consist of 80–90% cBN and a metallic W-Co or ceramic binder phase based on titanium and aluminium. Their grain size varies within a range of 0.5–10 μm . Low cBN-containing types are generally 45–65% cBN and have a ceramic binder phase based on titanium carbide or titanium nitride [Krat06]. They are as a rule very fine-grained (grain diameter < 2 μm). As Fig. 4.60 shows, the thermal conductivity and resistance of BN cutting tool materials against abrasive wear increases with

Table 4.1 PCBN types (BH) available on the market for cutting sintered steel and sintered forged steel, acc. to KRATZ [Krat06]

cBN content/ vol%	cBN grain diameter/ μm	Binder phase	Knoop hardness/GPa	Thermal conductivity/ (W/mK)	Bending strength/GPa
80	0.5–4	Co-W	26–30	80–100	1.4
85	0.5–5	Co-W	38–41	80–100	1.0
90	0.5–4	Co-W	40–43	100–110	1.3

Table 4.2 PCBN types (BL) available on the market for cutting hardened sintered steel and sintered forged steel in continuous cut, acc. to KRATZ [Krat06]

cBN content/ vol%	cBN-grain diameter/ μm	Binder phase	Knoop hardness/GPa	Thermal conductivity/ (W/mK)	Bending strength/GPa
45	0.5	Ti ceramic	24	34	0.6
50	2	Ti ceramic	28	44	0.6
60	1	Ti ceramic	33	45	1.2

increasing cBN content and grain size. Fine-grained BL types have a higher cutting edge quality, an essential prerequisite for producing high-quality surfaces in the hard finishing of components.

Tools made of cubic boron nitride are preferred for turning, drilling and milling hardened steel with a hardness of 55–68 HRC, high speed steel, grey cast iron, hard cast iron, sintered metals and high temperature alloys based on nickel and cobalt – materials which are very difficult to machine using cemented carbide tools. The main uses of BH types are generally roughing/pre-machining of hardened steel and

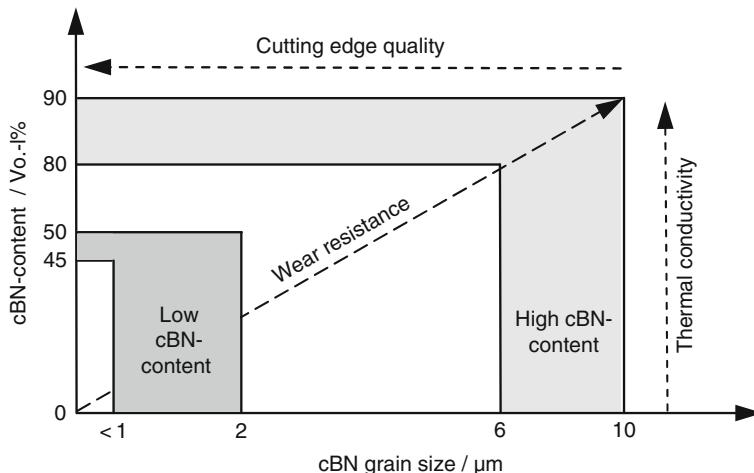


Fig. 4.60 Effect of cBN-content and grain size on properties of PCBN

iron materials, machining hard cast iron, sintered metals and all machining operations made on perlitic cast iron materials. Due to its relatively high thermal conductivity, they quickly divert the heat arising in the contact zone during roughing. BL types are primarily employed for hard finishing with depths of cut of < 0.5 mm (turning instead of grinding). They have increased edge stability and, for these machining tasks, better wear attributes than types containing higher levels of cBN (Fig. 4.61).

When machining hardened steels with BN cutting tool materials, tribochemical reactions between the cubic boron nitride (cBN) and components of the steel materials to be cut lead to intensified wearing of the boron nitride [Bömc89, Neis95,

Process: External cylindrical turning
 Material: 100Cr6 (60 HRC)
 Cooling: dry
 Cutting cond.: $v_c = 80$ m/min
 $f = 0.1$ mm
 $a_p = 0.5$ mm
 Cutting time: $t_c = 10$ min

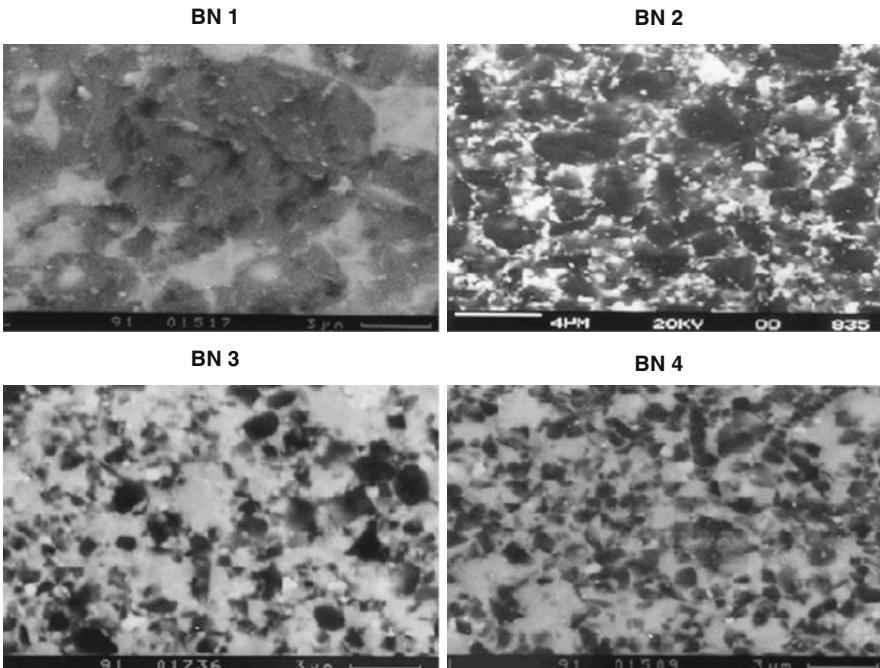
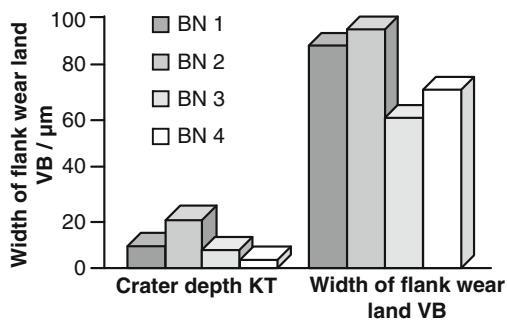


Fig. 4.61 Wear of different boron nitride cutting tool materials when turning 100Cr6

[Halp05](#), [Barr00](#), [Barr06](#)]. The higher the amount of cBN in the cutting tool material, the stronger this wear mechanism limits the performance of the BN cutting tool materials being used for hard machining.

In order to combat this wear phenomenon, cutting tool materials made of cBN are coated. Coating reduces tribochemical wear, increasing component quality, tool life and process safety [[Daws02](#), [Gey05](#), [Ohas04](#), [Okam05](#)]. Usually, cBN tools are coated with TiN, TiAlN or AlCrN after the PVD process. There are also solid BN plates coated with CVD- Al_2O_3 .

When machining materials of lower hardness (45–55 HRC), the higher Fe content in the materials leads to an increase in chemical wear. If hardened components must also be machined in the hard/soft transition phase, there is still another problem. Due to declining material strength, chip formation changes from lamellar to continuous chips. This is accompanied by changes in the friction conditions in the contact zone on the rake face, and thermal and chemical stress on the cutting edge increases. The BN cutting edge becomes very hot. In consequence, the solder with which the BN blank is soldered onto the cemented carbide base can melt and the insert can break. In these cases, the wear and performance attributes of cBN cutting tool materials can be significantly improved by coating them.

Turning tools made of polycrystalline cubic boron nitride are becoming more and more popular for finishing turbine blades composed of nickel-based alloys (see Sect. [7.6.5](#)). As opposed to cemented carbides ($v_c = 20\text{--}40 \text{ m/min}$), they can be employed at much higher cutting speeds ($v_c = 250\text{--}500 \text{ m/min}$) (Fig. [4.54](#)). In comparison to turning tools made of whisker-reinforced cutting ceramics or SiAlon, BN cutting tool materials are characterized by higher process safety, while the resultant component have a superior surface quality. When using the appropriate BN types (low-cBN-containing, binder phase TiC or TiN) and insert geometries (e.g. RNGX120700E), the types of wear that affect tool life as a rule are flank and rake face wear for high cutting speeds ($v_c > 250 \text{ m/min}$) and notch wear on the major and/or minor cutting edge for low cutting speeds ($v_c < 250 \text{ m/min}$). Besides low cutting speeds, small corner radii also are advantageous for the formation of notch wear. An emulsion is used as a cooling lubricant during the cutting process – dry machining is generally impossible.

4.6 Tool Designs

Various tool designs for separate processes have become available on the market, since different areas of use place particular demands on the design of the tools. In order to guarantee tool functionality, the following requirements should be considered already in the design phase:

- mechanical tool stresses (cutting forces),
- thermal tool stress (friction and deformation heat, coolant),
- rapid changing and secure positioning of the tool cutting edge,

- simple and fast replacement of worn tool parts,
- versatility in use,
- manufacturing and maintenance costs

Two main tool groups have emerged based on these requirements. Tools can be designed either as

- solid tools or as
- tools with inserts.

Included in the solid tool category are tools whose cutting part and shaft have been manufactured with the same material. In the case of tools with an insert, the shaft usually consists of a tougher and cheaper base material and the insert of a harder, more wear-resistant material (e.g. cemented carbide, cBN). Different designs have become established for different manufacturing processes. For example, in the case of milling and drilling, solid tools are more economical, whereas they are only used in special cases for turning.

In the following introduction to tools designs, special attention will be paid to turning tools. Further specifications regarding tool systems of other chip-removing machining processes with defined cutting edges can be found in [Chaps. 9 and 10](#).

4.6.1 Solid Tools

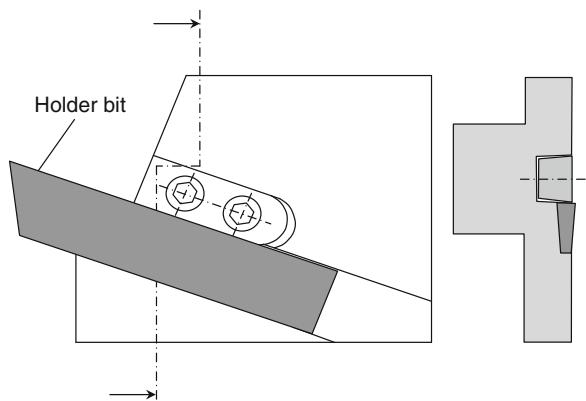
Solid tools are turning tools whose tool body and shaft consist of one material (e.g. high speed steel). This tool type, also called a tool holder bit, is fabricated by grinding the cutting part geometry on to a wide variety of basic shapes.

Parting tool holder bits already have a ready-made geometry. The angles required for the parting process are ground up to the tool orthogonal clearance. After grinding the tool orthogonal clearance, these tools are placed in a clamping as shown in Fig. 4.62. Final grinding occurs only on the flank face. Any necessary height correction is made by subsequent adjustment of the tool holder bit in the clamp.

One advantage of these tool holder bit is that theoretically freely selectable geometries can be ground onto them for every regrind. Figure 4.63 shows two forming tools in both round and flat design. The latter is designated as a tangential tool because the tool can be tangentially shifted, as opposed to the radial tool, which must be radially adjusted. Round tools are turned after regrounding by a corresponding angle. These types of forming tools are reground only on the rake face, since any regrounding of the flank face would alter the shape, which is very expensive to produce.

These tools are advantageous only for turning materials that do not require chip formers, since in these cases they can simply be reground. Round tools can be prepared until the circumference of the cutting part has been reduced to 25%.

Fig. 4.62 Parting off tool clamping



Solid tools are being used increasingly less commonly for turning, whereas usually tools with inserts are used. If soldering the cemented carbide inserts poses problems (e.g. in the case of small tools), turning tools can be fabricated completely out of cemented carbide. In order to guarantee a small amount of shaft deflection, in the case of boring bars for example, complete cemented carbide shafts are utilized, since cemented carbide has a larger elastic modulus.

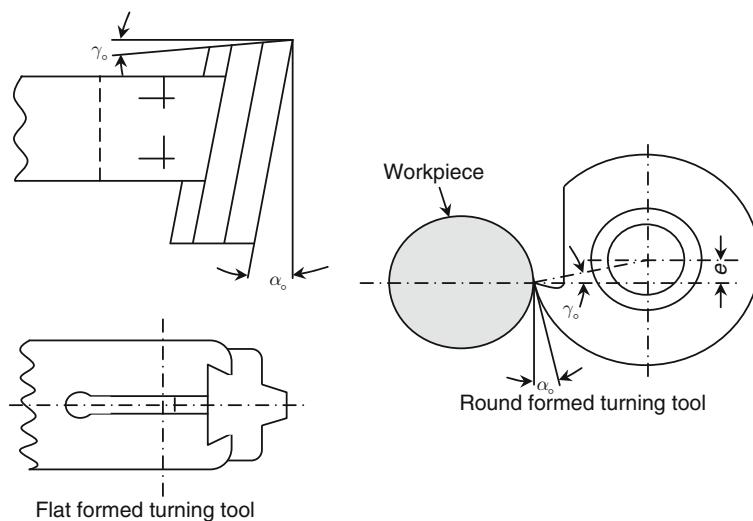


Fig. 4.63 Forming tools

4.6.2 Tools with Inserts

In accordance with joining techniques, tools with inserts should be classified as being either

- tools with fixed inserts (material connection) and
- tools with detachable inserts (frictional/positive connection).

4.6.2.1 Tools with Fixed Inserts

Soldering as a technique for joining the insert and tool shaft is now only used in exceptional cases for turning processes. It is more common in other areas, such as in milling and drilling processes or in the manufacture of woodworking tools. Today, inserts are most usually screwed or clamped.

For turning tool shafts, unalloyed construction steels with a strength of 700–800 N/mm² have proven successful. If tools for large chip cross-sectional areas are required, the strength should be increased to 800–1000 N/mm². Shaft cross sections are standardized in DIN 770-1 and DIN 770-2.

In the case of soldered tools as in Fig. 4.64, usually inserts made of PCBN or cemented carbide are soldered onto the tool shaft. High-speed steel is usually only

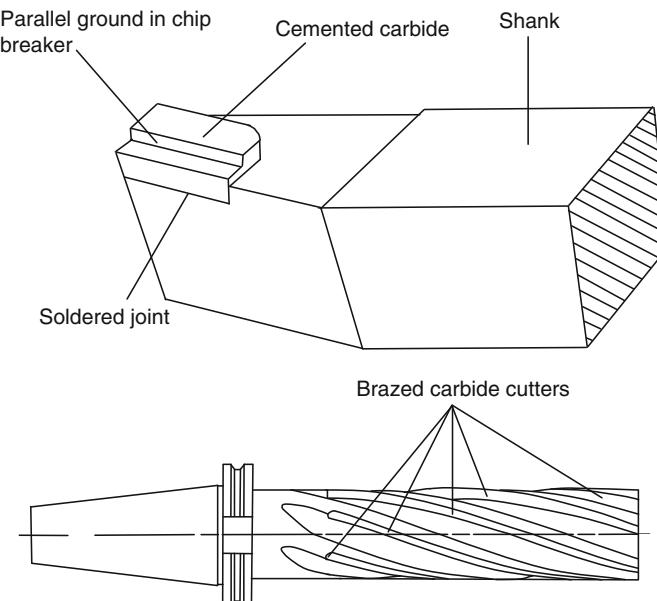


Fig. 4.64 Turning and milling tools with brazed carbide tipped cutters

soldered onto tools that have lower cutting edge requirements (e.g. in the case of saw blades for woodworking).

4.6.2.2 Tools with Detachable Inserts

As opposed to tool systems with soldered inserts, tools with clamped or screwed inserts have the advantage, among others, that several cutting edges can be used on a single insert. If a cutting edge has reached the end of its service life, the insert is turned or twisted after releasing the attachment, thereby bringing a new cutting edge into action. The term “indexable insert” comes from this process. The designation system for customary indexable inserts is standardized in DIN ISO 1832 and shown in Figs. 4.65 and 4.66.

On these tools, the inserts are attached with clamping devices on the tool holder. Of particular advantage is the safe and quick clamping of the inserts. In accordance with the selected tolerance class, insert replacement can be accomplished quickly during manufacture, since the cutting edge does not need to be positioned with respect to the workpiece.

The manufacturing tolerances of indexable inserts influence the production accuracy of the workpieces to a very large extent when there is a change of inserts (e.g. on NC machines). For inserts, a distinction is made between normal and precision design. In the case of a normal design, tolerances are in the range of ± 0.13 mm and for precision designs ± 0.025 mm, whereby workpiece tolerances of about ± 0.1 mm can be observed after an insert change without additional positioning.

According to the forms described in DIN ISO 1832, there is a variety of special designs that are non-standard inserts for machining processes with defined cutting edges.

Since clamping holders make it possible to clamp inserts made of diverse cutting tool materials, they can be adjusted well and quickly to the machining task at hand. Storage costs are relatively low, since these are limited essentially to storing the inserts and replacement parts for the tool holders [Corn74].

One must bear in mind that the components of different products are not interchangeable with regard to storage. For this reason, it is uneconomical to use different holder designs in one plant for the same purpose, since this makes the storage of replacement parts considerably more expensive.

There are now intelligent tool dispensing systems provided by different tool manufacturers that make it possible to manage the tools and to provide orders automatically adjusted to the purpose of the operator [Pitt04].

As shown in DIN ISO 1832, there are inserts with and without holes. Inserts without holes generally have a superimposed chip former and ones with holes have moulded chip formers. Inserts made of ceramic materials and PCBN are usually manufactured without a hole and can thus only be fixed in clamping holders and clamping fingers.

Clamping holders for turning are standardized in DIN 4983. Figure 4.67 shows the standard description for clamping holders by means of an example.

Example of insert notation:

C	N	M	G	12	04	08																																													
1	2	3	4	5	6	7	8	9	10																																													
1 – Insert form (ε_i)																																																						
A	85°	M	86°																																																			
B	82°	O																																																				
C	80°	P																																																				
D	55°	R																																																				
E	75°	S																																																				
H		T																																																				
K	55°	V	35°																																																			
L		W	80°																																																			
2 – Tool orthogonal clearance (α_0)																																																						
A		F																																																				
B		G																																																				
C		N																																																				
D		P																																																				
E																																																						
3 – Tolerance class																																																						
Acceptable variation in mm for																																																						
		<table border="1"> <thead> <tr> <th></th> <th>d / mm</th> <th>m / mm</th> <th>s / mm</th> </tr> </thead> <tbody> <tr> <td>A</td> <td>± 0.025</td> <td>± 0.005</td> <td>± 0.025</td> </tr> <tr> <td>C</td> <td>± 0.025</td> <td>± 0.013</td> <td>± 0.025</td> </tr> <tr> <td>E</td> <td>± 0.025</td> <td>± 0.025</td> <td>± 0.025</td> </tr> <tr> <td>F</td> <td>± 0.013</td> <td>± 0.005</td> <td>± 0.025</td> </tr> <tr> <td>G</td> <td>± 0.025</td> <td>± 0.025</td> <td>$\pm 0.05–0.13$</td> </tr> <tr> <td>H</td> <td>± 0.013</td> <td>± 0.013</td> <td>± 0.025</td> </tr> <tr> <td>J¹</td> <td>$\pm 0.05–0.15^2$</td> <td>± 0.005</td> <td>± 0.025</td> </tr> <tr> <td>K¹</td> <td>$\pm 0.05–0.15^2$</td> <td>± 0.013</td> <td>± 0.025</td> </tr> <tr> <td>L¹</td> <td>$\pm 0.05–0.15^2$</td> <td>± 0.013</td> <td>± 0.025</td> </tr> <tr> <td>M</td> <td>$\pm 0.05–0.15^2$</td> <td>$\pm 0.08–0.20^2$</td> <td>± 0.013</td> </tr> <tr> <td>N</td> <td>$\pm 0.05–0.15^2$</td> <td>$\pm 0.08–0.20^2$</td> <td>± 0.025</td> </tr> <tr> <td>U</td> <td>$\pm 0.05–0.15^2$</td> <td>$\pm 0.13–0.38^2$</td> <td>$\pm 0.05–0.13$</td> </tr> </tbody> </table>		d / mm	m / mm	s / mm	A	± 0.025	± 0.005	± 0.025	C	± 0.025	± 0.013	± 0.025	E	± 0.025	± 0.025	± 0.025	F	± 0.013	± 0.005	± 0.025	G	± 0.025	± 0.025	$\pm 0.05–0.13$	H	± 0.013	± 0.013	± 0.025	J ¹	$\pm 0.05–0.15^2$	± 0.005	± 0.025	K ¹	$\pm 0.05–0.15^2$	± 0.013	± 0.025	L ¹	$\pm 0.05–0.15^2$	± 0.013	± 0.025	M	$\pm 0.05–0.15^2$	$\pm 0.08–0.20^2$	± 0.013	N	$\pm 0.05–0.15^2$	$\pm 0.08–0.20^2$	± 0.025	U	$\pm 0.05–0.15^2$	$\pm 0.13–0.38^2$	$\pm 0.05–0.13$
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U	$\pm 0.05–0.15^2$	$\pm 0.13–0.38^2$	$\pm 0.05–0.13$																																																			
¹ Inserts with ground edges																																																						
² Depending on the insert size (See ISO-norm 1832)																																																						

Fig. 4.65 Standard description for inserts, acc. to DIN ISO 1832

Tool holders for inserts must fulfill the following basic tasks:

- Indexable inserts must always be clamped in the identical position after replacement.
- Position change due to cutting forces must be avoided.
- The support area must guarantee that the insert does not bend.
- The mounting must guarantee that heat arising in the process is diverted well into the tool holder.

4 – Fixation and/or chip breaker A : Top-left, two rectangular blocks. M : Top-middle, top block has a U-shaped cutout. R : Top-right, single rectangular block. B : Middle-left, two rectangular blocks. N : Middle-middle, single rectangular block. T : Middle-right, top block has a U-shaped cutout; angle $\beta = 40-60^\circ$. C : Bottom-left, two rectangular blocks. Q : Bottom-middle, single rectangular block. W : Bottom-right, top block has a U-shaped cutout; angle $\beta = 40-60^\circ$. H : Bottom-left, single rectangular block. X : Bottom-right, text: Drawing or accurate description required.			5 – Size of insert Insert sizes: I, I, I, I, I, d.																										
6 – Thickness of insert Thickness variations: s, s, s, s.			8 – Formation of cutting edge Formations: E (rounded), T (chamfered), F (sharp), S (chamfered + rounded).																										
7 – Design of cutting edge Radius r_e / mm: <table border="1"> <tr> <td>02</td> <td>0.2</td> <td>(R)</td> <td>00 for diameter in inches converted in mm.</td> </tr> <tr> <td>04</td> <td>0.4</td> <td></td> <td></td> </tr> <tr> <td>08</td> <td>0.8</td> <td></td> <td></td> </tr> <tr> <td>12</td> <td>1.2</td> <td></td> <td></td> </tr> <tr> <td>16</td> <td>1.6</td> <td></td> <td></td> </tr> <tr> <td>24</td> <td>2.4</td> <td></td> <td></td> </tr> </table> M0 diameter in metric system.			02	0.2	(R)	00 for diameter in inches converted in mm.	04	0.4			08	0.8			12	1.2			16	1.6			24	2.4			9 – Cutting direction Directions: right, left, neutral.		
02	0.2	(R)	00 for diameter in inches converted in mm.																										
04	0.4																												
08	0.8																												
12	1.2																												
16	1.6																												
24	2.4																												
10 – internal notations Condition of the edge, special chip former geometry, for example: P = polished W = with (expanded) finishing cutting edge																													

Fig. 4.66 Standard description for inserts, acc. to DIN ISO 1832 (continued)

- Cutting forces should be transferred to the holder such that the centering of the insert is supported.
- The use of a chip former is required depending on the machining task at hand.

Figure 4.68 shows different fixture designs for indexable inserts with (below) and without (above) a hole. The advantage of mountings for hole inserts is that all clamping elements in the holder are protected from chips. The indexable inserts

Notation example for tool holders:

Cartridge DIN 4985 –

C	S	G	N	R	32	25	M	16	Q
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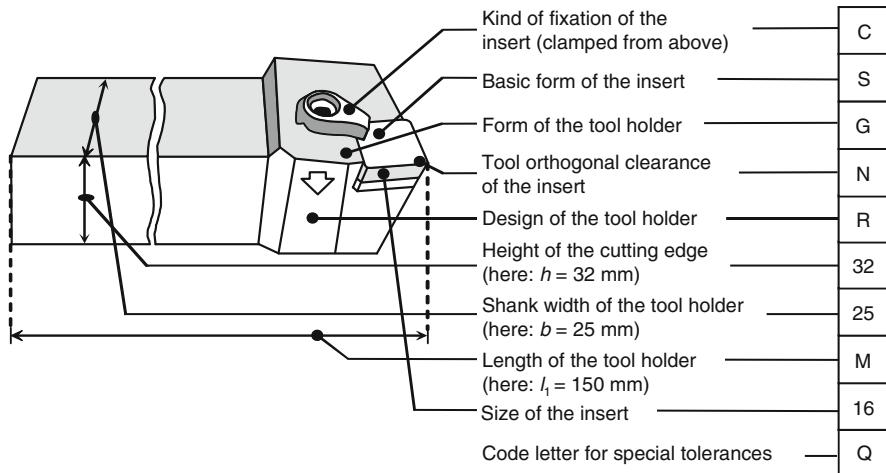


Fig. 4.67 Standard description for tool holders, acc. to DIN 4983

are unalterably centred and fixed in the insert seat, for example, by toggle levers or pins.

The simplest clamping is achieved with a clamping bolt. Another possibility is fixing with a clamping claw, which clamps the indexable insert and a continuously

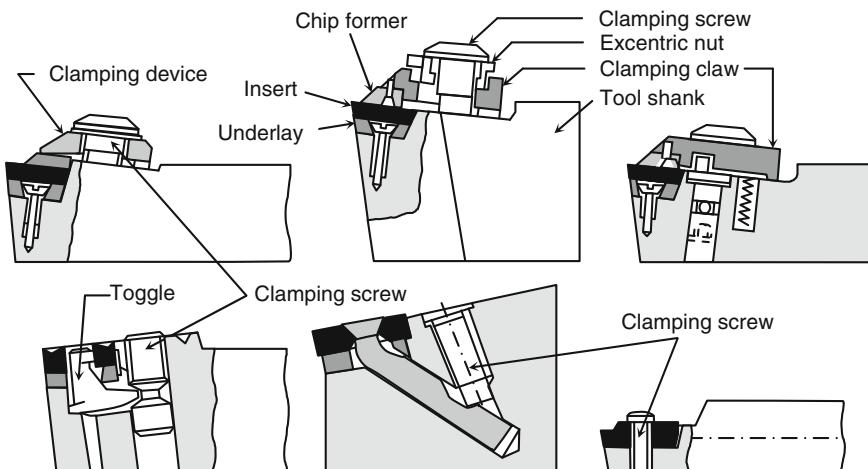


Fig. 4.68 Types of insert fixation (Source: Widia, Hertel)

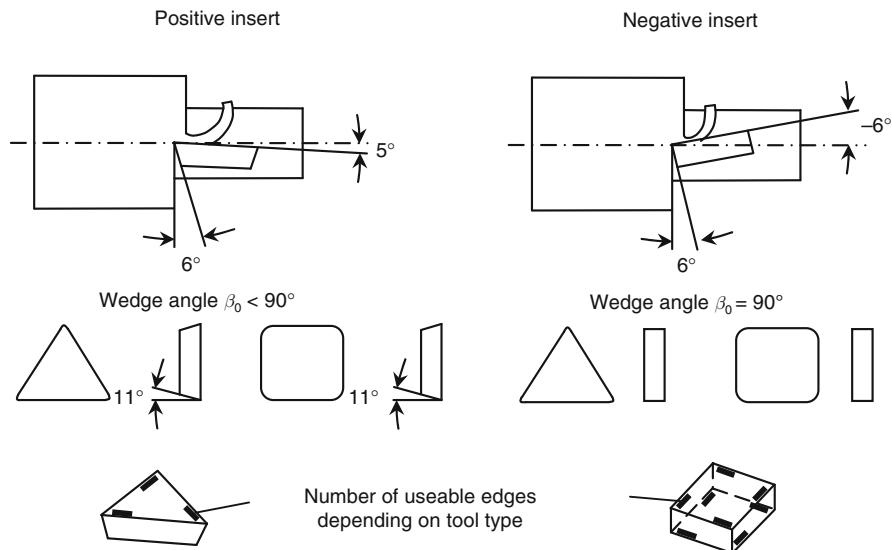


Fig. 4.69 Characteristic properties of positive and negative inserts

adjustable chip former. In this design, the chip former can be adjusted especially well to changing cutting conditions in order to guarantee a favourable chip form.

Tool holders with clamping fingers are used when the chip former is gradually adjustable and working conditions are mostly constant. If damage to the tool holder occurs, the individual parts can be replaced with replacement parts offered by the tool holder manufacturer.

In order to shorten setting-up and auxiliary process times in large-batch production, often special designs are used such as short tool holders or cassettes. Primarily for linked machines or tracer lathes, tools can be employed that can change used indexable inserts automatically during workpiece loading without loss of time.

There are negative and positive inserts. The criterion for this distinction is the size of the tool orthogonal rake angle when fixed, i.e. in the machining position. If there is a positive tool orthogonal rake angle, the insert is referred to as a positive insert and vice versa (Fig. 4.69).

Positive inserts have usable cutting edges only on the upper side. Indexable inserts for tool holders with incorporated positive tool orthogonal rake angles are provided with tool orthogonal clearances. If, as in Fig. 4.69, the tool orthogonal clearance of the insert is 11° (wedge angle $\beta_0 = 79^\circ$) and the tool orthogonal rake angle of the holder is $+5^\circ$, then the tool orthogonal clearance during tool engagement is $+6^\circ$. Negative indexable inserts have a wedge angle of 90° , making cutting edges available on both the upper and lower side of the insert.

Inserts with moulded or ground chip formers have the same basic shape as negative indexable inserts, but in effect they cut with a positive tool orthogonal rake angle

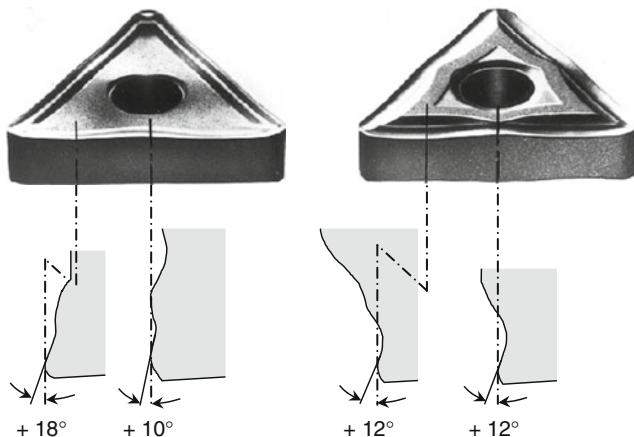


Fig. 4.70 Inserts with chip former (Source: Sandvik)

due to the geometry of the chip former (Fig. 4.70). This type of indexable insert thus combines the advantage of the positive tool orthogonal rake angle, a lower cutting force, with the higher number of cutting edges of the negative insert. The main function of the chip former is to produce short-breaking chips. On automated manufacturing devices, in which case process interference due to uncontrolled chip breakage is particularly disruptive, indexable inserts are therefore almost always provided with chip formers [Muel93].

One must bear in mind that moulded/ground chip formers should be adjusted to the respective cutting conditions [Sche78]. In order to expand the possible range of applications, often several chip breakers are arranged in a row. Small chip formers arranged near the corner radius guarantee good chip breakage in finishing operations.

Figure 4.71 provides examples of tools with detachable indexable inserts for drilling, milling, sawing and turning.

The large tool included angle of square inserts gives them a high level of cutting edge stability. As opposed to triangular inserts, they can only be used in form turning to a limited extent. Triangular indexable inserts have less cutting edge stability as a result of their small tool included angle. Very high surface qualities can be obtained by using round indexable inserts. The disadvantage of these however is that the smallest workpiece radius to be produced is prescribed by the cutting insert's geometry. Rhomboid inserts have been manufactured especially for copy turning operations. With these, deep and round contours can be reproduced.

4.7 Tool Preparation

Expensive tools, e.g. broaching tools, are as a rule prepared again, provided the cost to do so is lower than that of a new tool. Regrinding is almost exclusively done in

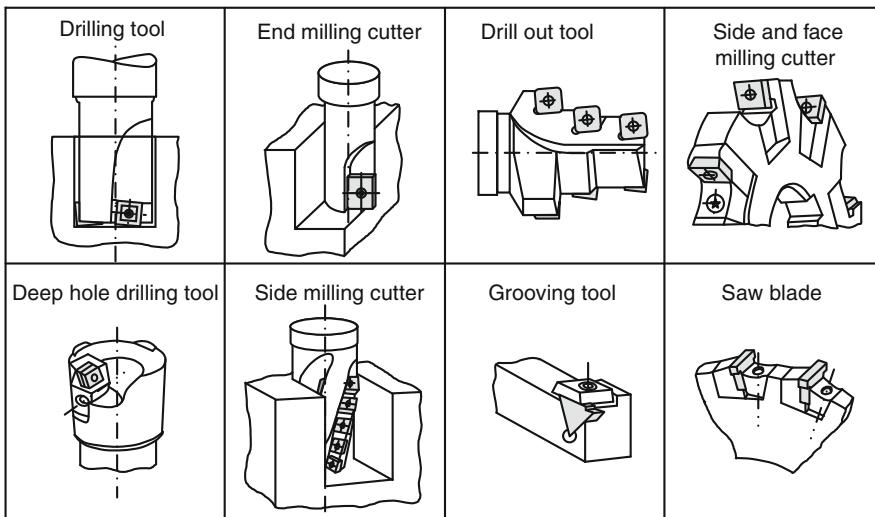


Fig. 4.71 Typical application examples

the case of special tools and tool designs with soldered inserts. Other tools that are typically prepared are solid carbide tools like drills and shaft millers, serration tools made of HSS and indexable inserts made of PKD or PCBN. Preparing indexable inserts made of cemented carbide or HSS is not economical because of the low material value.

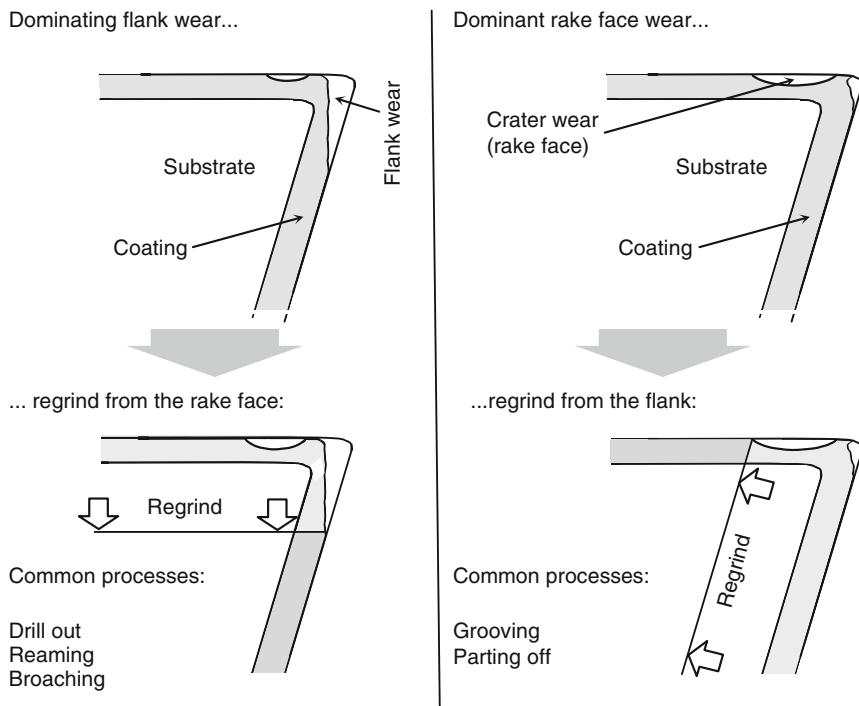
If a tool is to be prepared, regrinding must take place in a timely fashion in order to save on grinding and tool costs. This means that the cutting edge should not be used beyond a permissible level of wear. Figure 4.72 gives an overview of standard wear data used as tool life criteria for tools.

Grinding or regrounding tools is carried out as a rule on special 5-axis tool grinding machines. Only in this way we can always grind on the same angle and chip former geometries. Manually ground faces are imprecise and thus have a lot of influence on tool life and the chip forms produced.

When grinding high speed steel, the hardness and grain size of the grinding wheel must be adjusted to the high speed steel, depending on which grinding type is used, such as coarse or fine grinding. cBN or precious corundum discs are used as abrasives. Extensive information on the topic of abrasives and method variants of grinding can be found in [Kloc05a].

Special manufacturing processes in the area of finishing require sharply ground tool cutting edges. Before coating the tools, their surfaces are pretreated by irradiation. This produces a pronounced cutting edge radius on the tool. In order to be able to use sharply ground tools for finishing purposes nonetheless, different grinding strategies have been developed. Figure 4.73 shows different regrounding strategies corresponding to the type of wear arising. If crater wear is predominant, the flank face is reground so that the tools maintain the more wear resistant coating on the rake face. On the other hand, if flank face wear is predominant, the rake face is reground.

Cutting tool material	Measurand	Assumed wear data	
High speed steel	Width of flank wear land	VB	0.2 to 1.0 mm
	Crater depth	VB _{max}	0.35 to 1.0 mm
	Crater depth	KT	0.1 to 0.3 mm
Cemented carbide	Width of flank wear land	VB	0.3 to 0.5 mm
	Crater depth	VB _{max}	0.5 to 0.7 mm
	Crater depth	KT	0.1 to 0.2 mm
Cutting ceramics	Width of flank wear land	VB	0.15 to 0.3 mm
	Crater depth	KT	0.1 mm

Fig. 4.72 Guidelines for the tool life criterion**Fig. 4.73** Machined surfaces in regrounding

In regrinding, the grinding strategy and thus the surface quality has a large influence on the realizable tool life travel paths of the tool. An error-free regrind must therefore be guaranteed by a correct choice of abrasive and adjusted machining conditions. Moreover, one must take care during the grinding process that the tool does not become heated beyond annealing temperature in order to avoid rim zone damage. The avoidance of cracks caused by heat and force is of especial importance when preparing tools with soldered cutting inserts [Weir64].

If cemented carbide is used, regrinding, fine grinding and grinding chamfers and curves is done with diamond discs. Electrolytic grinding is also a very suitable method for pregrinding and finishing tools [Rein69].