Chapter 7
Tool Life Behaviour

The term tool life behaviour was introduced to describe material and cutting tool material behaviour during the machining process. The following definition applies:

Tool life behaviour is the ability of a working pair (tool and workpiece) to withstand a certain cutting process [DIN6583].

This is influenced by the cutting edge durability of the tool, by the machinability of the workpiece and by tool life conditions (Fig. 7.1).

Machinability is the property of a workpiece or material which allows chip removal under specified conditions. Cutting edge durability is the 4x ability of a tool to retain its cutting ability during machining. Cutting ability is the ability of a tool to machine a workpiece or a material under specified conditions [DIN6583].

Tool life behaviour is evaluated by means of the tool life conditions, criteria and parameters.

Machinability and cutting edge durability are both functions of the state variables force and temperature. These state variables are influenced in turn by the tool life conditions. The tool life conditions are all the conditions present during the cutting process or operational test. They comprise multiple components [DIN6583]:

- of the tool, e.g. its form, cutting edge geometry and cutting tool material
- of the workpiece, e.g. its shape and material
- of the machine tool, e.g. its static and dynamic stiffness
- of the cutting process, e.g. its kinematics and cutting edge engagement
- of the environment, e.g. the type of cutting fluid and thermal marginal conditions

In order to judge the tool life behaviour of the system encompassing the workpiece, tool, clamping, machine tool and coolant, tool life criteria are used which represent limiting values for undesired changes to the tool, workpiece or cutting process caused by machining. Examples of tool life criteria are:

- all measurable tool wear values, e.g. width of flank wear land,
- all measurable workpiece data, e.g. changes to roughness,
- all measurable values of the cutting process, e.g. changes to cutting power, chip temperature and form.

To describe the tool life of the system encompassing the workpiece, tool, clamping, machine tool and coolant, i.e. from the beginning of use until the achievement of

F. Klocke, *Manufacturing Processes 1*, RWTH edition,
DOI 10.1007/978-3-642-11979-8_7,

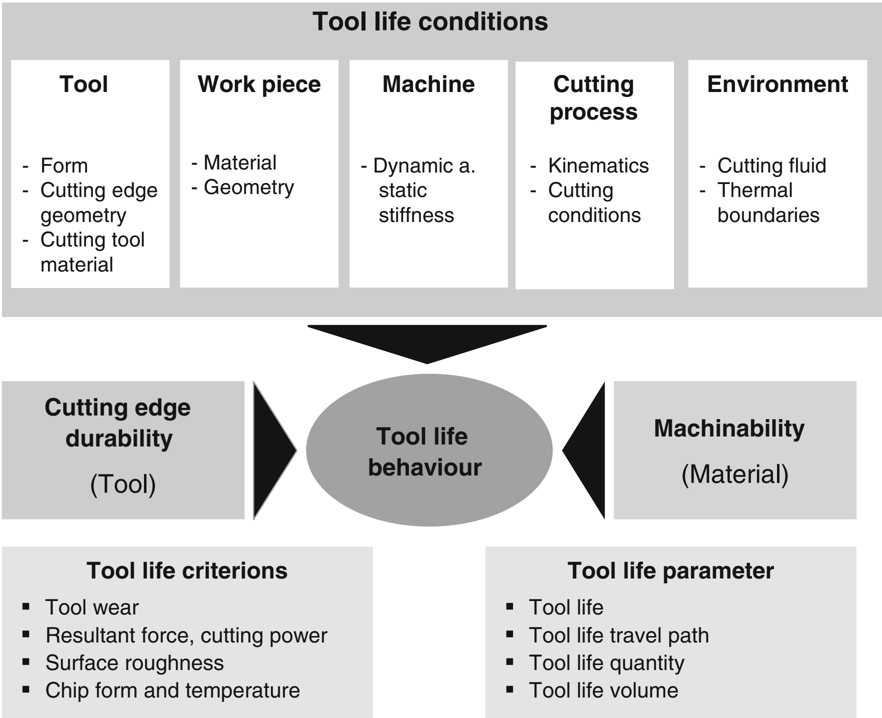


Fig. 7.1 Tool life behaviour

the tool life criterion under the influence of tool life conditions, tool life parameters are used. Tool life parameters are times, quantities or paths achieved in chipping under specified conditions until a tool life criterion is reached. These parameters include [DIN6583]:

- the tool life,
- the tool life travel path,
- the tool life volume
- and the tool life quantity.

In order to describe the tool life behaviour of the system of the workpiece, the tool, the clamping, the machine tool and the coolant in a clear way, the tool life conditions, criteria and parameters must always be specified. For example, when describing the tool life behaviour via tool life parameter, the tool life criterion and tool life condition are indicated in the index. If the tool life condition for describing machinability is selected, the tool life parameter and tool life criterion must be taken into account in the index. If the tool life criterion is used for description, the tool life parameter and condition must both be entered in the index. To conclude this section, the two following examples illustrate the relations explained above.

(a) description by means of the tool life condition:

$$v_{cT15;\,\mathrm{VB}\,0.2} = 200\,\frac{\mathrm{m}}{\mathrm{min}} \tag{7.1}$$

Specified parameters: $T = 15$ min (tool life parameter) and $\mathrm{VB} = 0.2$ mm (tool life criterion)

(b) description by means of the tool life criterion:

$$F_{\mathrm{cap}\,2;\,\mathrm{N}\,500} = 4000\,\mathrm{N} \tag{7.2}$$

Specified parameters: $a_{\mathrm{p}} = 2$ mm (tool life condition) and $N = 500$ (tool life parameter)

Due to the interaction between machinability and cutting edge durability, both parameters can be used to evaluate the machining system, assuming that a parameter is held constant with respect to the tool life behaviour (constant machinability for evaluating the cutting edge durability of different cutting tool materials given constant tool life conditions or constant cutting edge durability for evaluating the machinability of different materials given constant tool life conditions).

Tool wear is highly significant. In contrast to the time-varying state variables of the machining system, such as mechanical stress or temperature, tool wear can be defined relatively easily. The following sections will consider this more closely.

7.1 Determining Tool Life Parameters

Tests to determine tool life parameters can basically be executed using either slow testing methods or quick testing methods. Long-term cutting tests are executed for detailed descriptions of the stress of tools on machine tools. Such tests are costly in terms of both time and materials. As an alternative, different quick testing methods were developed in order to evaluate and compare the cutting edge durability and machinability of different materials while minimizing the required time and materials as much as possible.

7.2 Machinability

Parameters subjected to state changes during machining can be used as evaluation parameters for judging machinability. One must strictly define, however, whether the object of evaluation is the material (the workpiece) or the cutting tool material. This section will focus on the material, while the cutting tool material will be assumed to be constant.

The following parameters can be used to evaluate machinability:

- cutting force,
- tool life (or tool life travel path, quantity, etc.),

- the surface value of the workpiece and
- the chip form, etc.

It is often sufficient to use a single dominant parameter to evaluate machinability.

7.2.1 Tool Life

The tool life T_c of the tool is the most significant parameter for characterizing the machinability of a material. The tool life T_c is the time in min in which a tool performs from its first cut to its becoming unusable due to a specified tool life criterion under specified machining conditions.

7.2.1.1 Temperature Tool Life Rotation Test

The temperature tool life rotation test was developed for cutting tool materials with low temperature resistance (tool steels and high speed steels). The test is executed whenever not wear, but rather the influence of cutting temperature is the predominant factor for terminating tool life. The test is executed under constant tool life conditions until the cutting edge becomes unusable due to thermal conditions. This process of succumbing to thermal influences is also referred to as bright braking. Bright braking can be recognized by the formation of bright lines or lines of temper colour on the cut surface or on the machined workpiece surface or by the appearance of surface alterations. Altered chip forms and noises are also indicative of an advanced stage of damage to the cutting edge.

7.2.1.2 Wear Tool Life Rotation Test

The wear tool life rotation test is executed for cutting edge materials with a great temperature resistance (cemented carbide, cermet, ceramics, CBN). The test is executed whenever wear instead of cutting temperature is the predominant influence on tool life leading to the unusability of the tool. It is held using a longitudinal round cut with constant tool life conditions. After different cutting times, wear is measured on the flank and rake faces until the previously determined tool life criterion has been reached. It is generally sufficient to determine the width of flank wear land VB, the crater depth CD and the crater mean CM. The measurement results can be represented in a diagram (Fig. 7.2).

Using the wear curves in Fig. 7.2, respective value pairs can be formed for the tool life criterion from the cutting speed and cutting time which together form the tool life curve (Fig. 7.3).

The curve in Fig. 7.3 can be described approximately by means of a general exponential function. In a double logarithmic system, this function assumes the approximate form of a straight line (Fig. 7.4).

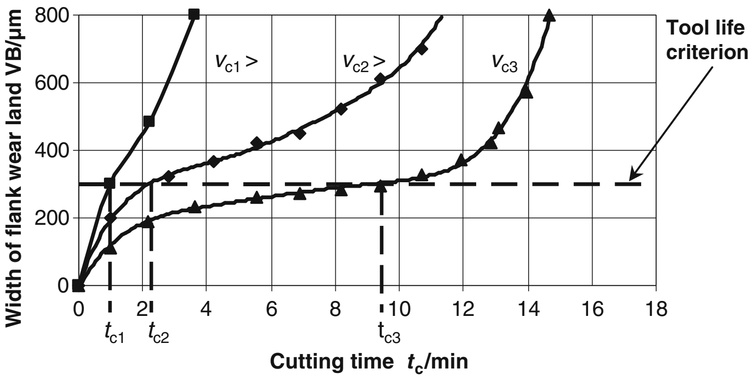
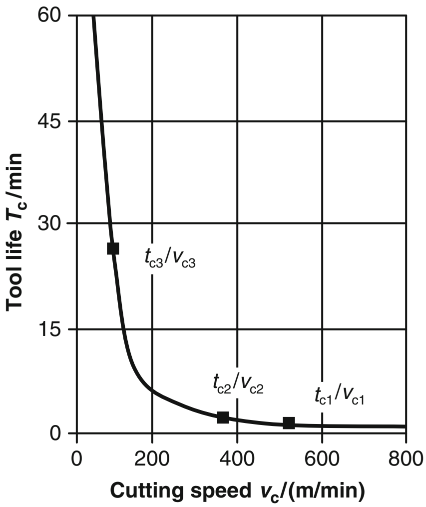


Fig. 7.2 Wear of uncoated carbide during machining heat treatable steel

Fig. 7.3 Tool life curve (heat treatable steel/carbide)



A straight line can generally be described by the linear equation:

$$y = m \cdot x + b \tag{7.3}$$

The following applies:

$$\log T_c = k \cdot \log v_c + \log C_v \tag{7.4}$$

And after taking the antilog, the following equation applies:

$$T_c = C_v \cdot v_c^k \tag{7.5}$$

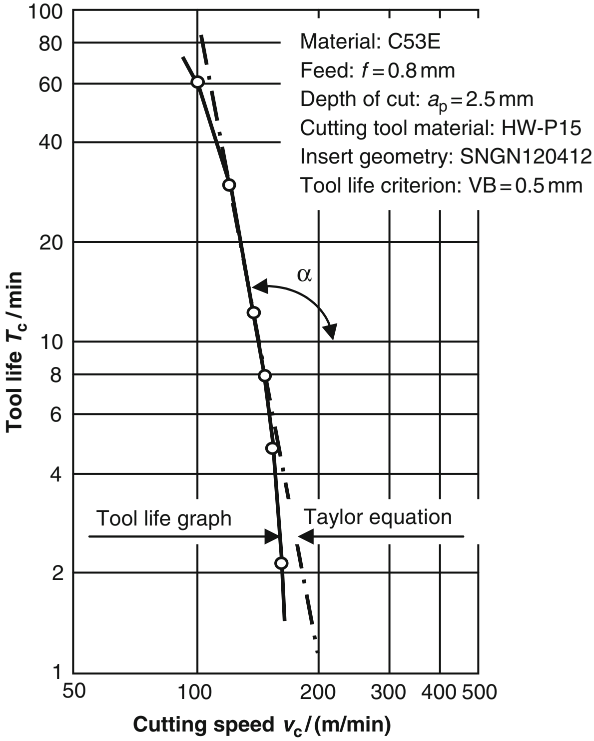


Fig. 7.4 Tool life curve in a logarithmical system (heat treatable steel/carbide)

This is referred to as the TAYLOR equation. The parameter C_v (ordinate intercept) indicates the tool life at a cutting speed of $v_c = 1$ m/min (standard tool life) and the parameter C_T (abscissa intercept) the cutting speed at $T_c = 1$ min (standard cutting speed). Thus the increase value k can also be expressed as follows:

$$\tan\alpha = k = -\frac{\log C_v}{\log C_T} \tag{7.6}$$

The tool life function in Eq. (7.4) is designated as a simple tool life function, since it only takes into account the influence of cutting speed on wear. This simple tool life function was developed by TAYLOR [Tayl07]. It is also known in an expanded form which takes the influence not only of cutting speed, but also of feed and cutting depth.

$$T_c = C \cdot v_c^k \cdot f_z^{k_{fz}} \cdot a_p^{k_a} \tag{7.7}$$

Research in to these basic relations essentially goes back to the American TAYLOR [Wall08]. Further research became necessary in his wake to make the tool life equations practically useful in a broad field of application. KRONENBERG was to exert the most influence on this research [Kron27]. Further research by KRONENBERG also showed that the application of similarity mechanics to chipping leads to the relations named in Eqs. (7.5) and (7.7). It must be generally borne in mind that the parameters in these functions are not constants. They can only be assumed to be approximately constant in certain areas.

7.2.2 Resultant Force

Knowledge of the magnitude and direction of the resultant force F or its components, the cutting force F_c, the feed force F_f and the passive force F_p, is a basis for

- constructing machine tools, i.e. designing frames, drives, tool systems, guideways etc. in line with requirements,
- determining cutting conditions in the work preparation phase,
- estimating the workpiece accuracy achievable under certain conditions (deformation of workpiece and machine),
- determining processes which occur at the locus of chip formation and explaining wear mechanisms.

Furthermore, the magnitude of the resultant force represents an evaluative standard for the machinability of a material, since greater forces tend to arise during the machining of materials which do not easily chip.

Resultant force is described by its amount and direction. In addition to amount, the force's effective direction can also have a significant effect on mechanically related changes to the tool or workpiece. Also, one can determine the stress state in the material in front of the workpiece cutting edge from the resultant force (Chap. 3).

The following will focus primarily on the influence of materials on the resultant force; geometrical and kinematic influences of the machining process will remain largely excluded from consideration. Exceptions to this will be noted where appropriate.

In practical applications, cutting force is often used instead of resultant force as an evaluation parameter. The cutting force is the component of the resultant force in the direction of primary motion. This procedure is reliable when the other components of the resultant force remain negligibly small. The specific resultant force or the specific cutting force may also be used as evaluation parameters (Chap. 3).

With respect to machinability tests in which the resultant force is used as an evaluation parameter, the distinction must be made between two different types of cut which respectively cause fundamentally different stress states in the material:

- Those that cause a biaxial stress state or one which lies at least in direct proximity to a biaxial stress state.
- Those that are far removed from a biaxial stress state.

7.2.2.1 Resultant Force Measurements

Measurements of resultant force are carried out by means of *dynameters*, which measure the average mechanical strain on the cutting tool in three directions which are orthogonal to each other [Schl29], preferably in the direction of the axes of the machine tool (Chaps. 3 and 8).

SALOMON discovered that there is an approximate exponential relationship between specific force and chip thickness [Salo26, Salo28]. Since this discovery, force measurements have been represented in relation to chip thickness values (Fig. 7.5). It must be borne in mind that extrapolations are not permissible, especially in the region of small chip thicknesses, since in this case at least the same exponential function is no longer valid.

By means of a representation in a double logarithmical diagram in which the exponential function follows a straight line, the specification parameters of the straight line may simply be determined by means of the axial sections and the gradient (see Fig. 7.6). The following equations apply:

$$\log k_z = \log k_{z1.1} + m_z \cdot \log h \tag{7.8}$$

$$k_z = k_{z1.1} \cdot h^{m_z} \tag{7.9}$$

$$m_z = \tan\alpha \tag{7.10}$$

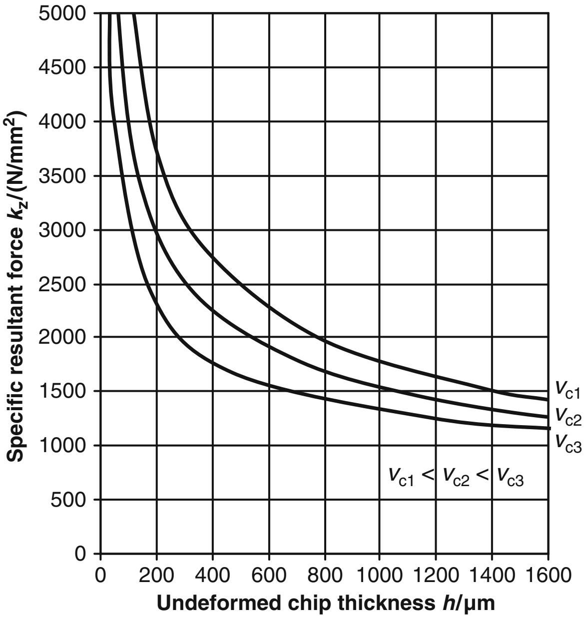


Fig. 7.5 Gradient of specific resultant force depending on the chip thickness (schematic diagram)

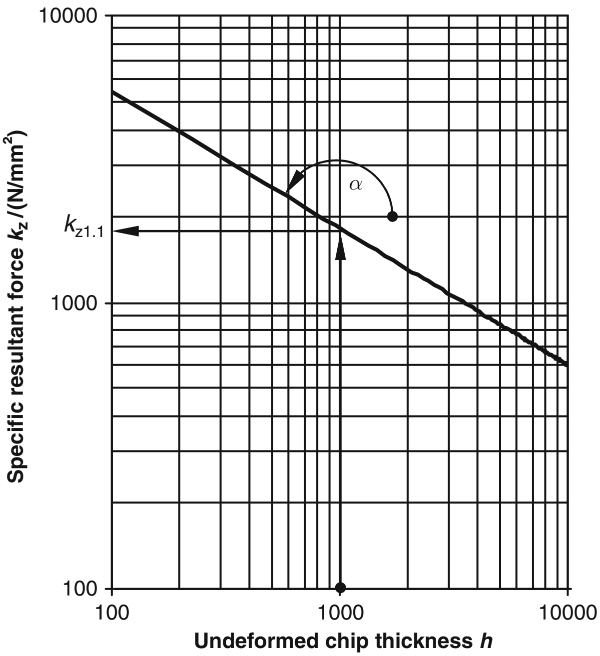


Fig. 7.6 Specific resultant force depending on the chip thickness in a double logarithmical diagram

7.2.3 Surface Quality

Surface quality can also be used to estimate machinability. The most important factors for this are the elastic and plastic deformations of the material in the area of the minor cutting edge [Djat52].

Low cutting speeds and certain material-tool combinations may lead to the adhesion of material particles on the rake face. This is referred to as the growth of built-up edges (Fig. 7.7). Due to mechanical and thermal stresses, the material which builds up on the rake face is sporadically stripped off und transferred to the workpiece surface.

Built-up edges are undesirable. They increase tool wear and lead to a poor surface quality (Fig. 7.7). With increased cutting speeds, this influence becomes increasingly insignificant.

The kinematic roughness is yielded by the relative motion between workpiece and tool and by the edge radius. During turning, it is primarily influenced by the form of the cutting edge and the feed. Figure 7.8 compares calculated and measured roughness values given a constant cutting speed without process disturbances caused by built-up edges.

BRAMMERTZ's theory accounts for plastic deformations and also the elastic spring-back of the material effected after the cutting edge reaches a certain area

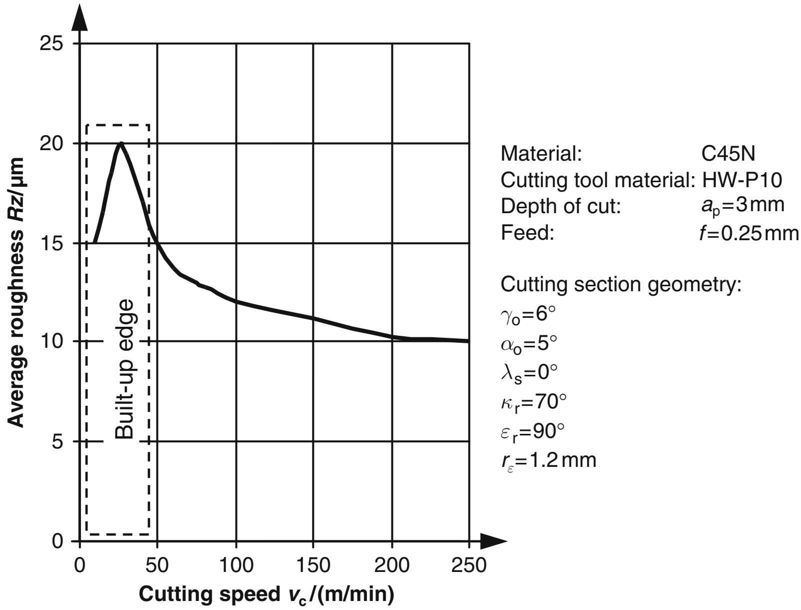


Fig. 7.7 Influence of cutting speed on surface quality

of the material [Bram61]. A minimum cutting depth must be achieved in order to guarantee chip formation. Otherwise, the material is only deformed elastically by the cutting edge. The yield point can be used to judge the elastic behaviour of the material to be machined.

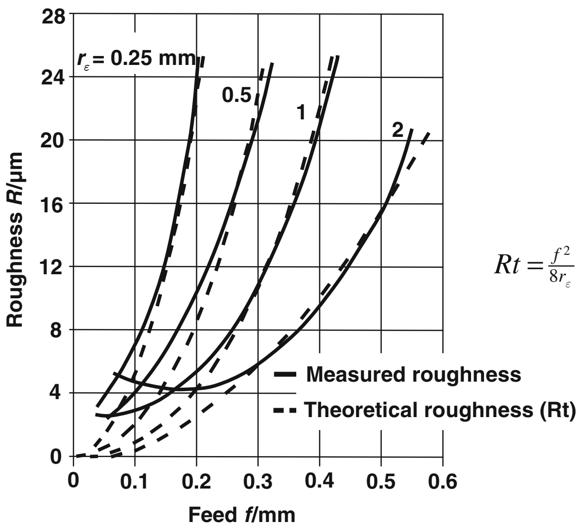


Fig. 7.8 Comparison of calculated and measured roughness

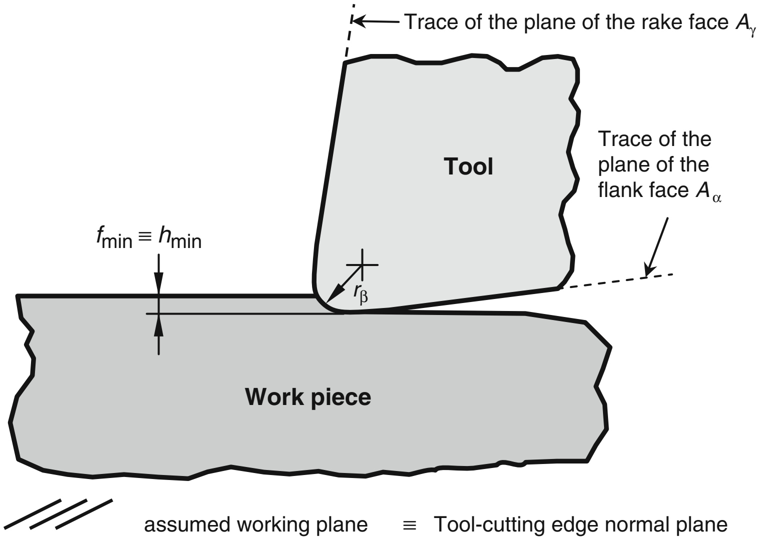


Fig. 7.9 Schematic sketch of minimum feed and minimum chip thickness

Since the material is not only stressed on the finished surface, but also on the cut surface, elastic deformation must also be taken into account in this case. This means that a minimum feed f_{min} must also be achieved as to allow a material separation to take place. This material separation depends on the cutting edge radius r_β and the yield point R_e of the material. In theoretical cutting theory, minimum feed is frequently projected to the tool-cutting edge normal plane P_n and referred to as minimum chip thickness h_{min} (Fig. 7.9).

There is no material separation below the minimum chip thickness value and thus no chip formation. As soon as the cutting edge radius assumes higher values than the chip thickness, the influence of the nominal tool orthogonal rake angle becomes insignificant, since the effective tool orthogonal rake angle becomes increasingly negative with an increasing cutting edge radius. The effective tool orthogonal rake angle represents a major influence on minimum chip thickness. If the minimum chip thickness is not reached, material accumulates in front of the cutting edge radius, as a result of which the workpiece material becomes pressed, squeezed and conveyed to the flank face (ploughing effect) [Albr60]. This compromises the achievable surface quality. The literature varies with respect to the minimum chip thickness h_{min}. As reference data for h_{min}, König and Klocke cite a factor of two to three of the cutting edge radius or the bevel width. In reference to the turning of steel, Sokolowski cites a dependence of minimum chip thickness on cutting speed, with h_{min} decreasing with increasing cutting speed [Soko55]. He investigated cutting speeds between 8 and 210 m/min, leading to values for h_{min}/r_n between 0.25 and 1.125.

Further notable influences on surface quality are material inhomogeneities and hardening.

7.2.4 Chip Form

In the machining of different materials, different chip forms are formed under the same tool life conditions. Examples of typical chip forms are shown in Fig. 7.10. Long chip forms make the evacuation of accumulating chips difficult. Flat helical chips tend to migrate outside the engagement length via the flank face, thus causing damage to the tool holder and the cutting edge. Ribbon, snarled and discontinuous chips represent an increased hazard to machine operators.

The formation of the different chip forms depends greatly on the friction conditions in the contact area between the chip and the rake face, the tool orthogonal rake angle, the cutting parameters and the material properties. Chip forms can be altered through alloying different chemical elements, such as phosphorous, sulphur and lead, or by means of a targeted heat treatment of the material. Chip breakage is generally favoured by the material's increasing strength and decreasing toughness.

Especially advantageous are chip forms that do not inhibit the machining process. These may not damage the tool system, the machine tool and the surface of the processed component. The spatial requirement for the chips and chip evacuation are also important parameters (chip volume ratio).

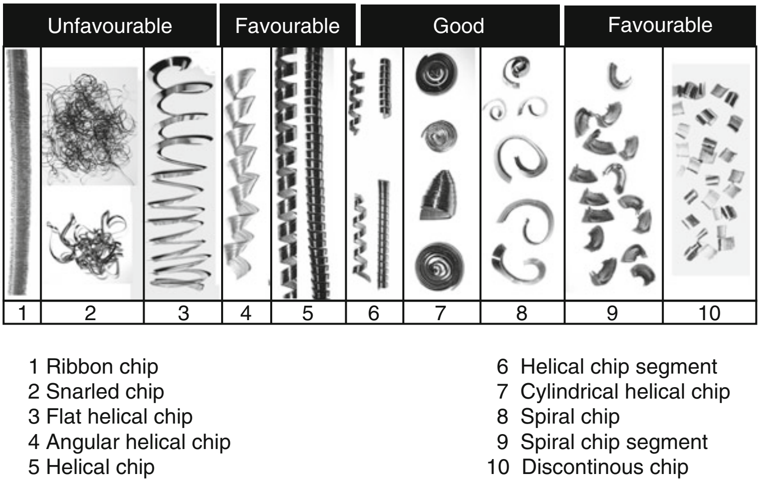


Fig. 7.10 Conventional chip forms

7.2.5 Cutting Speed

The following will evaluate the influence of cutting speed on machinability and discuss this influence via and analysis of specific cutting force. Figure 7.11 shows a typical curve for the specific cutting force when machining non-heat-treated steels.

Adhesion and the growth of built-up edges occur with lower cutting speeds. The adhesion tendency decreases with increasing cutting speed. A temperature range is

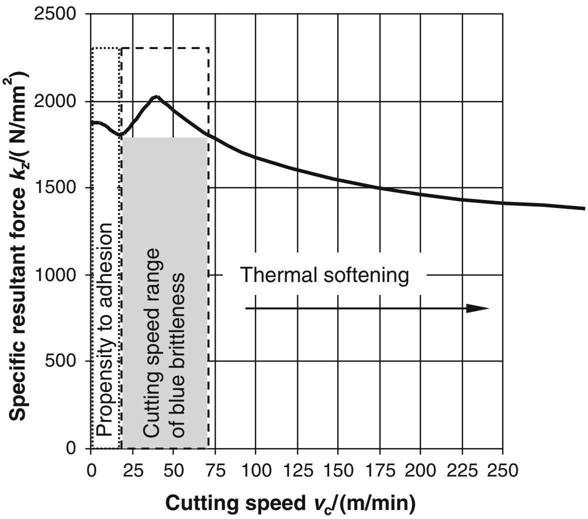


Fig. 7.11 Material properties depending on the specific resultant force

reached in which the specific resultant force again rises until a maximum value is achieved. This is the area in which blue brittleness occurs. Blue brittleness refers to a material condition in which the dislocation mobility is strongly limited by the interaction with nitrogen present in the microstructure. As a result, deformability decreases, thus increasing the specific resultant force. If the cutting speed is further increased, the temperatures in the shear zone increase sharply and the material is weakened, which causes the resultant force to be lowered again.

The following will describe materials frequently processed by chipping methods and their machinability.

7.3 The Machinability of Steel Materials

Steels can be classified according to their alloying elements, their metallographic constituents and their mechanical properties. Such a classification aids both in selecting a material with respect to the properties required for their future function and in determining machining conditions. Depending on their alloy content, steel materials are divided into the following groups:

- unalloyed steels,
- low-alloyed steels (alloy content $< 5\%$) and
- high-alloyed steels (alloy content $\geq 5\%$).

In the case of unalloyed steels, we must furthermore differentiate between those steel materials that are not to be heat-treated (common construction steel) and those that are (grade and special steel). Steels referred to as common construction steels (e.g. S235JR, S355J2G3) those whose mechanical properties exhibit minimum values. Construction steels are used in practice when no special requirements are made with respect to their structure.

7.3.1 Metallographic Constituents

The machinability of steels essentially depends on their respective crystalline structure. The crystalline structure of steel is mainly composed of the following components:

- ferrite,
- cementite,
- perlite,
- austenite,
- bainite and
- martensite

Depending on carbon content, alloying element content and heat treatment, one or more of these metallographic constituents predominate, whose mechanical properties (Table 7.1) affect the machinability of the given steel.

Table 7.1 Mechanical properties of metallographic constituents, acc. to VIEREGGE [Vier70]

	Hardness/HV 10	R_m/(N/mm^2)	$R_{p0,2}$/(N/mm^2)	Z/%
Ferrite	80–90	200–300	90–170	70–80
Cementite	>1100	–	–	–
Perlite	210	700	300–500	48
Austenite	180	530–750	300–400	50
Bainite	300–600	800–1100	–	–
Martensite	900	1380–3000	–	–

Ferrite (a-ferrite, cubic body-centred, maximal solubility for carbon: 0.02%) is characterized by relatively low strength and hardness and by a high deformability. Ferrite complicates the chipping process through the following:

- its strong tendency to adhere, which favours material smearing on the tool and the formation of built-up edges and built-up edge fragments.
- its tendency to form undesirable ribbon and snarled chips due to its high deformability
- poor surface qualities and increased burr formation on the workpieces

Cementite (iron carbide, Fe_3C) is hard and brittle and thus cannot be machined. Depending on carbon content and cooling speed, cementite can occur either by itself or as a structural constituent of perlite or bainite.

Perlite is a eutectoid phase mixture composed of ferrite and cementite. The eutectic point lies at 723°C and 0.83% carbon. Perlite appears in iron materials given a carbon content between 0.02 and 6.67%. Given a carbon content of 2.06%, it is the only metallographic constituent; above 2.06%, it is a constituent of ledeburite. For the most part, lamellar, linear cementite appears in perlite. After to a corresponding heat treatment (soft annealing) however, globular (spheroidal) cementite can also be formed. Heat treatable steels with a carbon content < 0.8% are referred to as sub-eutectoid.

During machining, the low deformability and greater hardness of perlite causes, on the one hand,

- significant abrasive wear and
- high resultant forces.

On the other hand,

- Perlite reduces the adhesion tendency and the formation of built-up edges,
- promotes the formation of favourable chip forms,
- causes less burr formation on the workpiece and
- improves the surface quality.

Austenite refers to the γ-mixed crystal of iron. It has a face-centred cubic structure. The maximum solubility for carbon is 2.06%. The structure exhibits only minimal hardness, although its strength can be increased by means of cold-forming. Austenite is the main constituent of many non-rusting steels and is not ferromagnetic. In unalloyed and low-alloy steels below approximately 723°C, austenite is transformed into perlite and, depending on the carbon content, into ferrite or cementite. Thus austenite is only present at room temperature in alloys. Examples of austenite builders are nickel (Ni), manganese (Mn) and nitrogen (N).

Essential properties of austenitic steel materials relevant to machining are:

- their high deformability and toughness: The high ductility of this material is based on the good plastic deformability of its face-centred cubic crystal lattice, which has four slip planes each with three slip directions and thus 12 slip systems in total. The result is a strong tendency of the material to form built-up edges, gluing-points and built-up edge fragments and to form unfavourable ribbon and snarled chips.
- their strong tendency to adhere to the cutting edge material: face-centred cubic materials tend to adhere much more than hexagonal or space-centred cubic metals, which means that built-up edges, gluing-points and built-up edge fragments occur more frequently when machining these steel materials.
- their tendency towards strain hardening: in the area of the workpiece surface being created, the deformation of the material caused in chip formation causes a hardening of the material forming the surface. This results in additional stress to the tool cutting edge, especially in subsequent cuts.
- their relatively low heat conductivity, which is 1/3 lower than in unalloyed steels. It impairs heat removal through the chip and increases the thermal stress of the cutting edge.

Bainite forms in the temperature range between those of perlite and martensite: iron diffusion is no longer possible, and carbon diffusion is already considerably hindered. There are two main forms of bainite:

- acicular bainite (with continuous cooling and isothermal transformation)
- granular bainite (only with continuous cooling)

Independently of the form, bainite consists of carbon-supersaturated ferrite, with the carbon partially precipitated as carbides (e.g. Fe_3C) whose size (from rough to very fine) is determined by the conversion temperature. Among the acicular bainite forms, one distinguishes according to the transformation temperature between lower bainite (strong similarity to martensite) and upper bainite (strong similarity to perlite).

Martensite forms when a steel material with a carbon content of > 0.2% is rapidly cooled from the austenite temperature range to a temperature below the martensite starting temperature. Due to the rapid cooling, the carbon dissolved in austenite is forced to remain dissolved in the mixed crystal. By means of a diffusionless transformation, a tetragonally deformed, body-centred martensite lattice forms from the face-centred cubic austenite lattice.

Martensite has a fine-acicular, very hard and brittle microstructure which is difficult to machine. The cutting tools used are subject to

- an increased abrasive wear and
- high mechanical and thermal stresses

7.3.2 Carbon Content

The metallographic constituents of steels can be seen in the iron-carbon phase diagram (Fig. 7.12). Steels with C-content < 0.8% precipitate ferrite when cooling from the austenite zone. The remaining austenite disintegrates into perlite below 723°C. With a carbon content of 0.8%, only perlite is formed as a eutectoid mixture of ferrite and cementite. In steels with a C-content > 0.8%, perlite and secondary cementite are formed. The secondary cementite is precipitated from the austenite predominately at the grain boundaries.

The machinability of steels with a carbon content < 0.25% is essentially characterized by the properties of the free ferrite. Because of the high deformability of the material, the resulting cut surface roughness is high. Since ferrite has an intrinsically large adhesion tendency, built-up edges form with low cutting speeds. Tool wear and cutting temperature only increase slowly with a rising cutting speed. Because of ferrite's low strength, only tools with the greatest possible positive tool orthogonal rake angle (e.g. with turning: $\gamma_o > 6°$) should be used. In order to decrease the adhesion tendency and to improve the surface quality, oils are generally used as cutting fluids, with their lubrication properties being more significant than their cooling effect. Steels with a carbon content of < 0.25% pose particular problems for grooving and parting off, as well as for boring, reaming and thread die cutting.

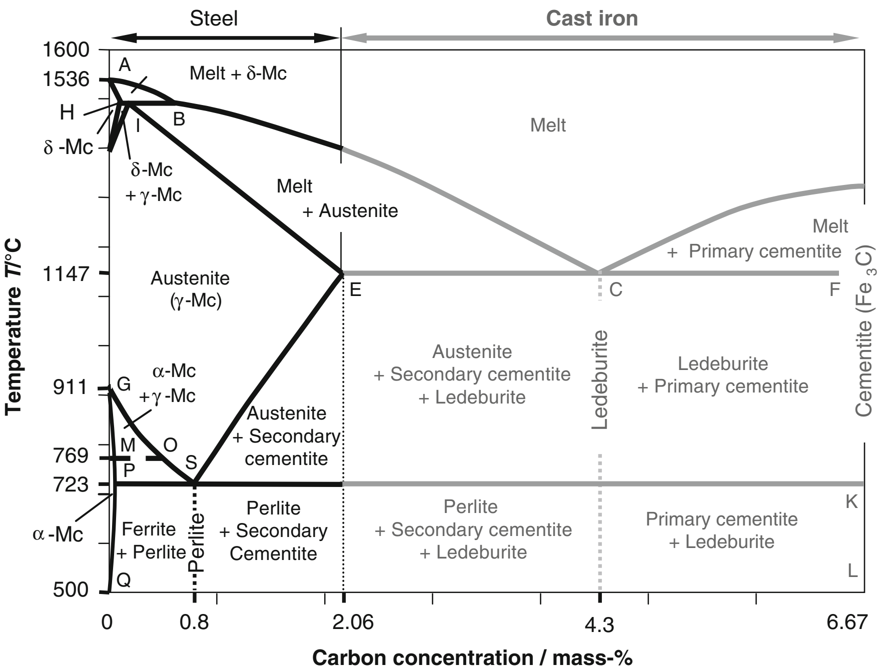


Fig. 7.12 Section of the iron-carbon phase diagram (metastable system)

Due to high deformability and relatively low cutting speeds, the achievable surface qualities are markedly inferior. Moreover, burr formation occurs to a greater extent.

The amount of perlite in the structure increases with increasing carbon content (0.25–0.4%). In this way, the special property of this metallographic constituent is granted a stronger influence on the machinability of the material. The strength of the structure increases, its deformability decreases. This results in the following:

- a lower adhesion tendency and thus a shift of built-up edges to lower cutting speeds,
- increasing tool wear and higher temperatures in the tool contact zones due to the higher material strength,
- an increase in abrasive wear,
- an improvement of surface quality and chip form.

Machinability can be improved by means of coarse grain annealing given a low C-content and by means of normalizing given a C-content above 0.35%. Tools with a positive orthogonal tool rake angle should be used. Cold-forming has a positive effect on machinability, especially with respect to chip formation. This is important to the extent that these materials are frequently deformed through cold extrusion and then finished through cutting.

A further increase of carbon content (0.4–0.8%) causes a further decrease of the amount of ferrite with a corresponding perlite increase until only perlite remains at 0.83% carbon. Due to increased strengths, high rake face temperatures already arise at low cutting speeds. At the same time, the increasing mechanical stress of the rake face causes increased wear, also in the form of crater wear. Problems related to chip form are more rare. The chip form improves with increasing carbon content, with wear also increasing. Steels with a carbon content of 0.4–0.8% are generally regarded as having good machinability only with respect to surface quality and chip form.

After slow air cooling, the structure of supereutectoid heat-treated steels ($C > 0.8\%$) consists of secondary cementite and perlite. Perlite formation initiates directly from the austenite grain boundaries. Secondary cementite precipitates at the grain boundaries given a carbon content significantly exceeding 0.8%. The free cementite forms shells around the austenite/perlite grains [Schu04]. Such shells cause very strong wear in cutting processes. In addition to the strongly abrasive effect of the hard and brittle metallographic constituents, the high pressures and temperatures which arise cause an extra stress for the cutting edge. Strong crater and flank face wear is to be observed even at relatively low cutting speeds. The cutting edge parts must have a stable design. For turning, tools with positive orthogonal tool angles γ_o up to 6° and mildly negative tilt angles λ_s up to –4° should be used.

7.3.3 Alloying Elements

Alloying and trace elements can influence the machinability of steel by changing the composition or by forming lubricating or abrasive inclusions. The following will discuss the influence of some important alloying elements on the machinability of steel.

7.3.3.1 Manganese

Manganese improves temperability and increases the strength of steel (approx. 100 N/mm^2 per 1% alloying element). Because of its high affinity to sulphur, manganese forms sulphides with sulphur. Manganese contents up to 1.5% facilitate machinability in steels with low amounts of carbon due to good chip formation. In the case of steels with larger amounts of carbon, however, machinability is negatively affected by increased tool wear.

7.3.3.2 Chrome, Molybdenum, Tungsten

Chrome and molybdenum improve temperability, thereby influencing the machinability of case-hardened and heat-treated steels in terms of structure and strength. In the case of steels with a greater carbon or alloy content, these elements and also tungsten form hard special and mixed carbides which impair machinability.

7.3.3.3 Nickel

Nickel is among the group of elements which expand the γ-phase zone in iron alloys. By adding nickel, the strength of steel materials increases. Nickel increases toughness, especially at low temperatures. This generally leads, especially in the case of austenitic nickel steels (with larger amounts of nickel), to an unfavourable machinability.

7.3.3.4 Silicon

Silicon increases the strength of ferrite constituents in steel. With oxygen, it forms, in the absence of stronger deoxidation agents like aluminium, hard Si-oxide (silicate) inclusions. This results in increased tool wear.

7.3.3.5 Phosphorous

Alloying phosphorus, which is carried out only in some free-cutting steels, leads to segregations in the steel which cannot even be removed with subsequent heat treatment and heat deformations and to an embrittlement of the ferrite. This allows for the formation of short-breaking chips. Up to an amount of 0.1%, phosphorous has a positive effect on machinability. While higher amounts of phosphorous lead to an improvement of the surface quality, they also cause increased tool wear.

7.3.3.6 Titanium, Vanadium

Titanium and vanadium, even in small amounts, can increase strength considerably due to the extremely dispersed carbide and carbon nitride precipitations. They lead to a high grain refinement, which has a negative effect on machinability with respect to mechanical stress and chip form.

7.3.3.7 Sulphur

Sulphur is only slightly soluble in iron, but it forms, depending on the alloying components of the steel, various stable sulphides. Iron sulphides (FeS) are undesirable, as they exhibit a low melting point and deposit primarily at the grain boundaries. This leads to the unwanted “red brittleness” of steel. What are desirable on the other hand are manganese sulphides (MnS), which have a much higher melting point. The positive effects of MnS on machinability are short-breaking chips, improved surface quality and a smaller tendency towards forming built-up edges. With an increased inclusion length, MnS exerts a negative influence on mechanical properties like strength, strain, area reduction and the impact value, especially when it is included transversely to the strain direction. In practice, however, special alloying additives (e.g. tellurium, selenium) can effectively help to counteract the deformability-related extension of MnS.

7.3.3.8 Lead

Lead is not soluble in iron; it is present in the form of submicroscopic inclusions. Because of the low melting point, a protective lead film forms between the tool and the material. This film reduces tool wear. The mechanical stress of the tool can be lowered by up to 50%, with the chips becoming short breaking. This effect is exploited especially in machining steels (see Sect. 7.4.1).

7.3.3.9 Non-metallic Inclusions

The elements added to the steels for deoxidation, i.e. aluminium, silicon, manganese or calcium, bind the oxygen released by steel solidification. The hard, non-deformable inclusions then found in the steel, e.g. as aluminium oxide and silicon oxide, diminish machinability, especially when the oxides exist in the steel in larger amounts or in linear form [Wink83]. However, by choosing a suitable deoxidatizing agent, the machinability of steel can also be positively influenced. For example, under certain machining conditions, wear-inhibiting oxidic and sulphidic layers may form after deoxidation with calcium-silicon or ferro-silicon [Köni65, Opit67].

One measure for improving the machinability of steels deoxidized with aluminium is calcium treatment. In this process, calcium is added to the steel melt by means of secondary metallurgy. This converts the sharp-edged aluminium oxides present in steel in conventional melting which act abrasively during cutting into globular and, in machining conditions, plastifyable calcium aluminates. Analogously to manganese sulphides among machining steels, these calcium aluminates are able under certain cutting conditions to form friction-reducing and/or wear-inhibiting coatings in the contact zones of the tool cutting edge [Köni65, Töns89, Kloc98a, Zink99].

7.3.4 Types of Heat Treatment

When executed in a targeted manner, heat treatments can influence microstructures with respect to quantity, form and the configuration of their constituents. In this way, mechanical properties and thus machinability can be customized to the given requirements. According to DIN EN 10052, heat treatment is defined as follows: Heat treatment is a sequence of heat treatment steps during which a workpiece is in whole or in part subjected to time/temperature sequences in order to cause a change of its properties or its structure. In certain circumstances, the chemical composition of the material may be changed during the treatment (thermochemical treatment).

Three basic categories of heat treatment may be distinguished:

- establishing an even structure in the entire cross-section, which to a great extent is in a state of thermodynamic equilibrium (e.g. soft annealing structures, structures consisting of ferrite or austenite) or in thermodynamic disequilibrium (e.g. perlite, bainite, martensite),

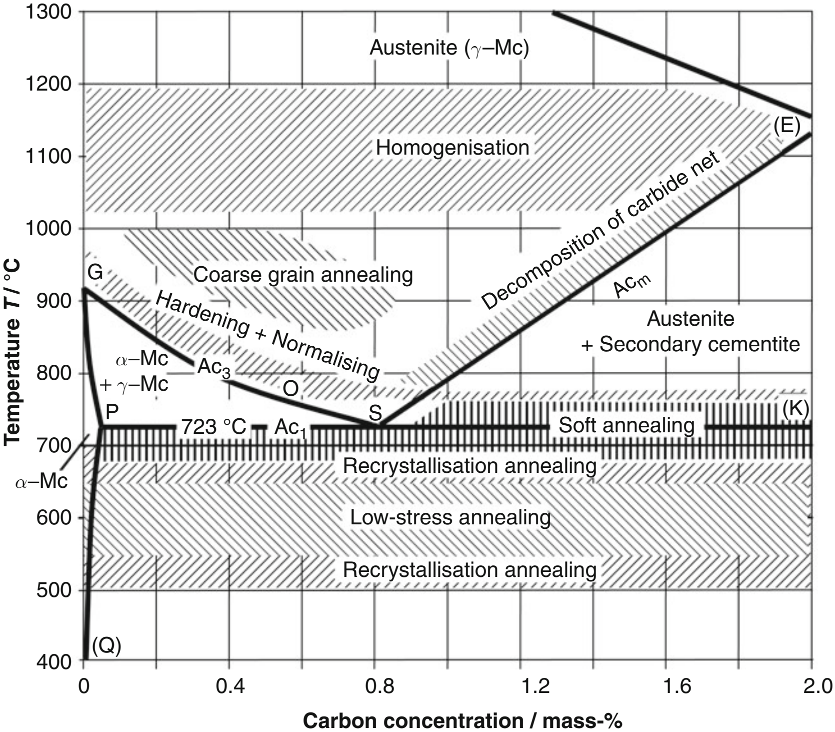


Fig. 7.13 Partial iron-carbon diagram with heat treatment range data

- establishing a hardening structure that is restricted to smaller areas of the cross-section at unaltered chemical composition (in particular: surface layer tempering),
- establishing structure types that are starkly differentiated over the cross-section, especially in the rim zone, as a result of a change of the chemical composition (nitriding, carbonitriding, carburizing, carburizing hardening).

There are different heat treatment methods with a broad field of application which, depending on the chemical composition of the steel, can affect machinability in a targeted way with respect to, for example, chip form and tool wear. The temperature ranges for the individual types of heat treatment can be found in Fig. 7.13.

7.3.4.1 Homogenization

Homogenization is defined according to DIN EN 10052 as follows: annealing at a high temperature with a hold sufficiently long enough to minimize local differences in chemical composition due to diffusion-related segregations.

7.3.4.2 Coarse Grain Annealing

Coarse grain annealing is defined according to DIN EN 10052 as follows: annealing at a temperature usually considerably above Ac_3 with a hold sufficiently long enough to achieve coarse grain.

Coarse grain annealing followed by isothermal transformation is utilized for subeutectoid steels with a C-content of 0.3–0.4% (ferritic-perlitic steel) in order to produce a coarse-grain structure with a ferrite network which is as closed as possible in which either perlite or bainite is enclosed [Hors85, Schu04]. Tool wear when machining such a structure is relatively low, and chip formation generally good. High surface qualities remain achievable. The application of coarse grain annealing to improve machinability is, however, limited by interference by strength properties and due to financial considerations.

7.3.4.3 Normalizing

Normalizing is defined according to DIN EN 10052 as follows: a heat treatment comprising austenitizing and subsequent cooling in still air.

Through normalizing (+N), a nearly even, fine-grain structure is achieved whose machinability is, depending on the carbon-content, determined by the predominant structural constituent, i.e. either by ferrite (low wear, poor chip formation) or perlite (increased wear, improved chip formation) [Hors85]. A α/γ transformation takes place in the process. Subeutectoid steels are heated to temperatures above Ac_3.

Supereutectoid steels are heated to temperatures 30–50°C above Ac_1 or, if the carbide network is to be dissolved in structures with higher carbon content, to temperatures exceeding Ac_3. In general, this heat treatment should be suited to achieving an even, fine-grain perlitic-cementitic structure.

Stronger carbide bands cannot be fully dissolved, which means that normalized supereutectoid steels can still cause relatively high tool wear. However, they allow a high surface quality of the finished surfaces.

7.3.4.4 Soft Annealing

Soft annealing is defined according to DIN EN 10052 as follows: a heat treatment for reducing the hardness of a material to a specified value.

Soft annealing (+A) is applied in order to take away the high hardness and low deformability from structures with lamellar perlite or lamellar perlite and cementite. By means of annealing at temperatures just below the PSK line (Fig. 7.13) – or, if necessary, oscillating at the PSK line – and subsequent slow cooling, a soft and low-stress state can be created as the lamellar perlite and strands of cementite are caused to disintegrate. The goal is to achieve a perlite structure which is as granular as possible consisting of ferrite with globular cementite. Such a structure is soft and easily deformable. The machinability of such a structure becomes more favourable with respect to wear effects on the tool, while chip formation worsens to the extent that ferrite predominates in the structure.

7.3.4.5 Annealing to Specific Properties

In practice, other types of treatment are required in addition to classic soft annealing which target specific properties or structures.

A special type of soft annealing is annealing to spherical carbides (+AC) or spherical cementite (also referred to as GKZ annealing). The goal of this type of annealing is to mould the cementite completely as a sphere. Temperatures are held in the region of the PSK line for a longer period of time, if necessary oscillating at this temperature. Since, when the cementite is fully moulded, its machinability approximates that of pure ferrite, it may worsen [Opit64].

Given steels with a carbon content of 0.10–0.35%, materials with Widmannstätten structure may be created by means of high austenitizing temperatures, long holding times and rapid cooling. The result is an acicular ferrite with extraordinarily finely distributed lamellar cementite. Such a structure is characterized by good chip formation and chip form, though it exhibits poor usage properties [Schu04].

Case-hardened steels are heat treated to a ferrite/perlite microstructure (+FP). In this state, they can achieve a similarly good machinability as with machining steels with low carbon content, and this with respect both to low tool wear and to good chip formation. A further heat treatment for improving machinability used primarily with case-hardened and heat-treatable steels is annealing to a specific tensile strength (+TH).

For energy-saving reasons, targeted heat treatments are executed directly using the heat of the forge, e.g. controlled cooling from the heat of the forge (BY annealing) [Fasc80]. Machinability investigations [Wink83] have shown that heat-treatable steels cooled from the heat of the forge (e.g. C45E+BY) may exhibit a more favourable wear behaviour than the same materials in a heat-treated or normalized state. Differences with respect to chip formation could not be ascertained. The reason lies in the relatively coarse-grain structure of the steels used and in the fact that the ferrite network also encloses the perlite grains during the shear process.

7.3.4.6 Recrystallization Annealing

Recrystallization annealing is defined according to DIN EN 10052 as follows: a heat treatment with the goal of achieving the formation of new grain in a cold-formed workpiece by means of nucleation and growth without a phase change.

Recrystallization annealing refers to annealing following cold forming at a temperature below Ac_1, which for steel is usually between 500 and 700°C, without causing a α-γ-transformation of the crystal lattice. This is usually applied between the individual shaping stages, e.g. when cold rolling or cold drawing metal sheets and wires [Schu04, Meta06].

Strain hardening becomes noticeable in the structure through the occurrence of displacements, glide lines and the splitting of brittle crystal types (cementite).

Recrystallization annealing prevents changes to properties associated with this, such as increased hardness and strength and decreased strain and toughness. If the material reaches its forming limit in the process of cold forming, a recrystallization must be executed to form new grains. The composition of the microstructure is not newly formed during recrystallization; only the grains are newly formed. The greater the strain is, the greater the tendency towards the formation of new grains. At high strain levels, the size of the newly formed grains can lie below that of the original grains [Schu04].

7.3.4.7 Low-Stress Annealing

Low-stress annealing is defined according to DIN EN 10052 as follows: a heat treatment comprising heating and holding the material at a sufficiently high temperature and a subsequent cooling appropriate for eliminating internal stresses as much as possible without essentially changing the structure.

The usual temperatures for low-stress annealing steel workpieces lie between 550 and 650°C. A microstructural transformation does not take place.

Low-stress annealing is mainly applied to workpieces which exhibit high internal stresses either as a result of irregular cooling following casting, welding, forging or another thermal process or after strong mechanical processing through milling, turning, planing, deep drawing, etc. Low-stress annealing serves to reduce these stresses. This prevents the release of existing internal stresses during the further processing of such workpieces and the formation of geometrical deviations resulting from warpage. The strain hardening in the deformation zones induced by cold-forming, however, is reduced by means of recrystallization annealing.

7.3.4.8 Hardening

Hardening is defined according to DIN EN 10052 as follows: a heat treatment comprising austenitizing and cooling under conditions conducive to an increase in hardness caused by the more or less complete transformation of austenite into martensite and, if applicable, bainite.

When hardening steel (+Q), the precipitation of carbon from the γ-mixed crystal which happens at a normal cooling speed counteracted through a high cooling speed. At supercritical cooling speeds, martensite forms after falling short of the Ms-temperature (Ms = martensite starting) [Schu04].

At cooling speeds lower than the critical cooling speed, the conversion processes proceed in the intermediate stage and in the perlite stage [MPI61, MPI72, MPI73]. Transformation in the intermediate stage is basically characterized by that fact that only the carbon can diffuse. Processes of continuous and isothermal transformation for the heat-treatable steel C45E are shown in Figs. 7.14 and 7.15.

These structure types are not as easy to machine because of their greater strength. Chip formation can be considered good. The cutting edge must have a stable design.

7.3.4.9 Heat Treatment

Heat treatment is defined according to DIN EN 10052 as follows: Hardening and tempering at higher temperatures in order to achieve the desired combination of mechanical properties, especially high toughness and ductility.

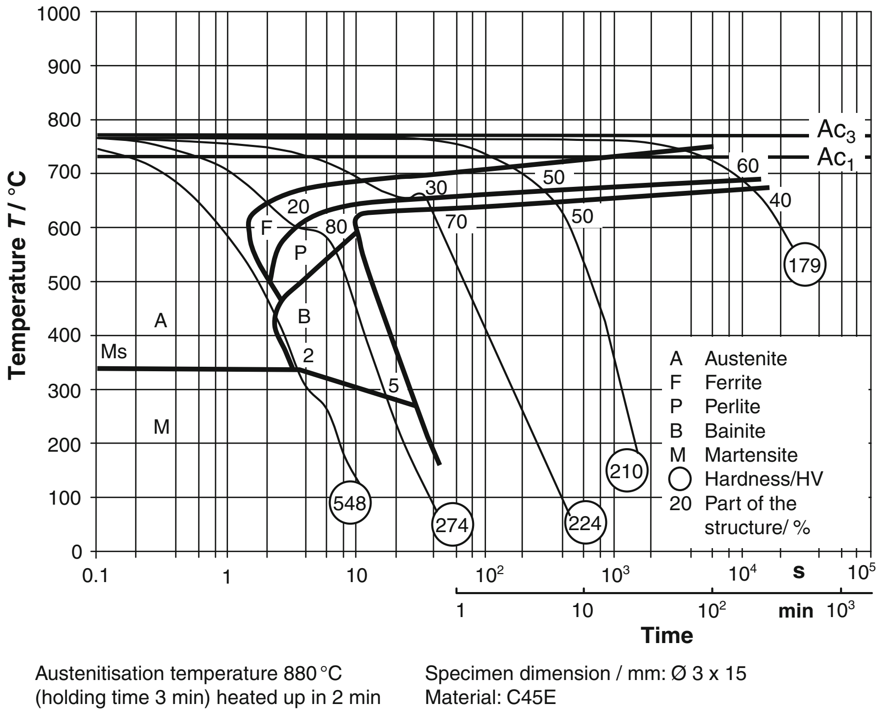


Fig. 7.14 Time-temperature-transformation diagram for continuous cooling of C45E steel, acc. to MPI

The strength values of steel can also be increased through quenching and tempering (+QT). When the material is tempered, the martensite formed during hardening is broken down again in a targeted way through re-heating. At low tempering temperatures, carbon precipitates in a finely distributed form, while at higher temperatures coarser cementite grains develop [Schu04]. The machinability of the tempered structure increases with increasing martensite decay.

Several possibilities for a targeted influence on structure by various heat treatments are shown in Fig. 7.15 using the heat-treatable steel C45E as an example.

The allocation of the partial images to the heat treatments is as follows (Fig. 7.16):

1. *Coarse grain annealing.* Metallographic constituents: coarse-grain perlite with lamellar cementite, a ferrite network between the grains (white).
2. *Normalizing.* Metallographic constituents: perlite with lamellar cementite, ferrite. These are the same constituents as in coarse-grain annealing, but the structure is finer-grain and more homogeneous.
3. *Quenching and tempering.* Structure: tempered martensite.
4. *Soft-annealing*. Metallographic constituents: ferrite (white) with globularly shaped cementite.

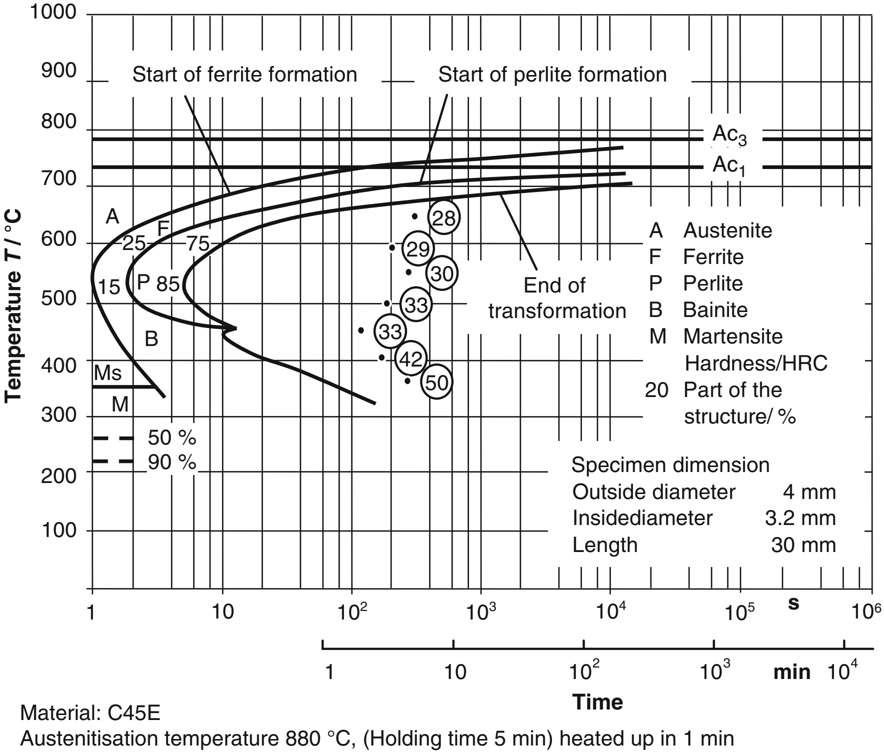


Fig. 7.15 Time-temperature-transformation diagram for isotherm conversion of C45E steel, acc. to MPI

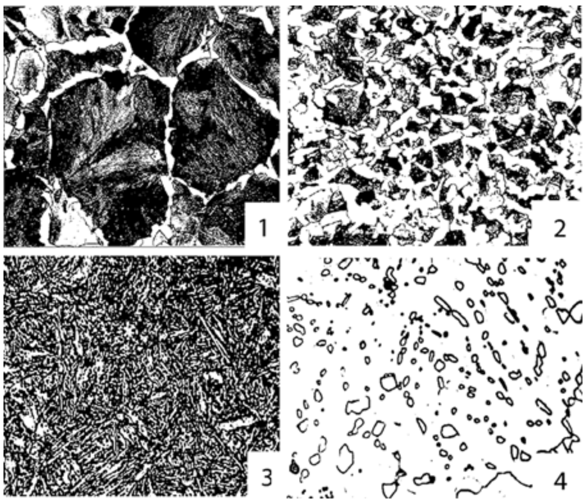


Fig. 7.16 Crystalline structure for different heat treatments (C45E)

7.4 Machinability of Various Steel Materials

In addition to the categorization of steels according to their alloy content, they are also categorized in practice with a view to their uses and applications. The categories are:

- free cutting steels,
- case-hardened steels,
- heat-treatable steels,
- nitrided steels,
- tool steels and
- rust- and acid-proof, heat-proof and highly heat resisting steels.

7.4.1 Machining Steels

The free cutting steels are materials which are distinguished by especially good machinability. Characteristics in this respect are a favourable breakage, clean workpiece surfaces and low tool wear. They can be machined without difficulty on automatic lathes and multi-spindle lathes and are thus highly suited to serial and mass production. The high cutting speeds can be used, these being often limited by the workpiece diameter and the respective machine tools (multi-spindle lathes). Depending on their chemical composition and field of application, free cutting steels are subjected to heat treatments like carburizing hardening or quenching and tempering.

These favourable properties with respect to machinability are given to machining steels primarily by adding different alloying elements. The main alloying elements, sulphur, lead and phosphorous, as well as tellurium, bismuth and antimony reduce the strength of the material in the shear zone. The reduced strength results in short-breaking chips with a low level of compression.

Machining steels are predominately cut with coated cemented carbide tools and moulding tools consisting of high speed steel. The development of wear on these tools is relatively slow. The low-carbon machining steels in widespread use (e.g. 11SMn30, 11SMnPb30, 9S20) are characterized by high ferrite and low perlite content. The result of this is a low level of abrasive tool wear. Even at low cutting speeds, adhesion in the form of built-up edges plays an important role with respect to tool wear. When machining low-carbon machining steels at low cutting speeds ($v_c < 100$ m/min), friction-reducing layers (MnS, Pb) form which contribute to reducing tool wear [Zink99].

The machinability of free cutting steels is relatively low for machining purposes. Reasons for this are the large amount of ferrite in the microstructure as well as the strength-reducing properties of the main alloying elements.

The achievable surface qualities are considerably influenced by the adhesion tendency of the material caused by the formation of built-up edges. MnS layers, Pb and P reduce this adhesion tendency, positively influencing the surface quality of the workpieces.

As a function of the steel production process, phosphorous causes a formation of segregations (demixing) in the steel because of its relatively small diffusion coefficient. This undesirable property can only be partially removed through subsequent heat treatment. High-temperature diffusion annealing can only compensate microsegregations (crystallization segregations, concentration differences). In contrast, segregations at the macro-scale (macrosegregations) cannot be removed extensively by means of a subsequent heat treatment. This is because of the large distance from the middle to the edge of the steel ingot. Sufficiently eliminating macrosegregations would require annealing times of such a length as would lead to a significant coarsening of the grains. Smaller macrosegregations can be achieved by deoxidizing the steel prior to casting.

Nitrogen and carbon precipitations are promoted by phosphorous and induce an embrittlement of the α-mixed crystal (ferrite embrittlement). These precipitations increase with increasing temperature, which means that tempering brittleness occurs to a greater degree and the impact value is already reduced from 100°C. The degradation of strength properties caused by phosphorous are responsible for the embrittlement of machining steels at the average shear level temperature of approximately 200–400°C. The result is a more favourable chip formation with short-breaking chips when machining. The adhesion tendency in the contact zone continues to be reduced and the surface quality is positively influenced. Machining steels contain up to 0.1% phosphorous.

The positive effect of MnS on machinability is explained by the fact that both the internal friction of the material in the shear zone and the friction in the contact zone are reduced. Recent investigations prove the inclusion of MnS as the basis for glide line formation in the shear zone. Since sulphur reduces tool wear but hardly influences tempering brittleness, it is favoured over phosphorous as alloying element.

MnS causes the chips to be short, improves the surface quality of the workpiece and reduces the tendency toward the development of built-up edges. The form of sulfidic inclusions in machining steel is determined by the quantity of oxygen released upon solidification. Three different types of sulphides are found in steel:

- Type 1: This type of sulphide forms as a liquid phase at an oxygen content of > 0.02% corresponding to the quasiternary system Fe-MnO-MnS in iron-rich melts. After solidification, this sulphide appears in the form of regularly distributed, globular particles or irregularly rounded particles. These brittle type 1 sulphides are clearly separated from each other in the steel, enclosed in cells.
- Type 2: In iron melts with an oxygen content of < 0.01%, this form of sulphide precipitates from melts rich in manganese sulphide at the primary grain boundaries as a MnS phase similar to an eutectic point. Type 2 sulphides are not present in steel in enclosed cells, but rather grow as if radiating from individual centres. In the process, furcations form or, in some places, the sulphides growing out of the funnel-shaped cells fuse together.
- Type 3: This type of sulphide crystallizes from Fe melts with a reduced melting point. An essential requirement for the formation of these sulphides is a carbon

and silicon content between 0.1 and 0.4% and an aluminium content between 0.05 and 0.3%. Type 3 sulphides precipitate in interdendritic regions. However, like those of type 1, they are evenly distributed. Their composition and form correspond to that of the angular, face-centred cubic α-MnS.

With respect to machinability, type 1 sulphides are considered to be the most favourable. As a result, different metallurgical measures are taken to create this sulphide type for free cutting steels. The formation of round manganese sulphides is favoured, for example, by alloying tellurium. The machinability of machining steel improves in an essential way with an increasing quantity of larger manganese sulphides. Such sulphides prevent the pressure welding of the ferrite grains by forming scale-like layers and create a protective zone between the chip and the tool [Köni66].

Machinability is improved by adding up to 0.35% lead. Lead does not dissolve in α-Fe and penetrates the steel microstructure in the form of sub-microscopic inclusions. The strength properties and toughness of steels are negatively influenced, especially in the region of 250–400°C. Lead liquefies at relatively low temperatures ($T_S = 326°C$). When machining steels containing lead, a thin lead film may wet the contact surfaces between the tool and the workpiece. This reduces the tendency towards pressure welding and facilitates the shearing-off process. At cutting speeds exceeding 100 m/min or with large feeds, the lead film becomes ineffective and wear accelerates. The specific cutting forces fall by up to 50% and the chips become short-breaking.

Adding lead (about 0.25%) to machining steels can increase tool life by approximately 50–70%. The effect of lead on tool wear depends on the cutting speed. Given a lead-content increase of 0–0.29% and cutting speeds below 100 m/min, there is a reduction of flank face wear on HSS tools. In the cutting speed region above 100 m/min, an increased lead content has a negative effect on the development of flank face wear. With respect to crater wear, the effect of lead is not affected by the cutting speed.

As opposed to sulphur, lead can be alloyed with almost all steels, but these steels may only be stressed in a temperature range below 200°C. Typically, about 0.15–0.30% Pb is added to machining steels. Substituting lead as an alloying element in machining steels gains in importance with a view to its harmfulness to health [Wink83]. Similarly to lead, tellurium, bismuth and antimony also help to achieve improved chip breakage and bring about a lubricating effect in the contact zone which reduces tool wear. Investigations have shown that adding bismuth considerably improves the machinability of lead-free machining steels (e.g. 16SMn30) with respect to their chip formation, surface roughness and forces arising during machining. In comparison with alloys containing lead, however, the onset of abrasive wear on uncoated cemented carbide tools is considerably more aggressive when turning. This effect is not to be observed with coated cemented carbide tools. Machining steels in which lead has been replaced by tin exhibit a significantly reduced tendency towards chip breakage and are thus only recommendable for methods with interrupted cutting, such as milling [Esse06].

Free cutting steels in common use include 9SMn28, 9SMnPb28, 35S20, 45S20.

7.4.2 Case-Hardened Steels

The case-hardened steels include unalloyed machining steels, grade and special steels as well as alloyed special steels. Common to all of these is a relatively low carbon content ($C < 0.2\%$). Case-hardened steels are predominately utilized in the manufacture of wear-stressed and variably stressed parts like cogwheels, gear shafts, joints, connectors etc.

Case-hardened steels are almost exclusively processed by chip removal prior to case-hardening. Since the microstructure of these steels contains mostly ferrite and only a small amount of perlite, the onset of wear on the tool is low. Beyond this, cutting speeds should be kept above 200 m/min in order to avoid the growth of built-up edges. The machining of case-hardened steels is mostly achieved with coated cemented carbide tools belonging to application group P (e.g. HC-P10) or with cermets in order to withstand the thermal stress caused when machining at high cutting speeds.

Because of the high ferrite content and low perlite content in their microstructure, the machinability of case-hardened steels in a soft state is low.

The adhesion tendency of case-hardened steels leads to the formation of built-up edges and to bad surface qualities when procedures with low cutting speeds are used, such as tapping, boring, broaching and shaping. In order to improve machinability with respect to surface roughness, case-hardened steels are heat treated to a certain ferrite/perlite structure (BG) or to a certain strength (BF), depending on their alloying elements. Coarse-grain annealing is often applied to alloyed case-hardened steels in order to reduce difficulties caused by their adhesion tendency. At the same time, this reduces the strong tendency of the structure towards linearity, which is highly disadvantageous for machining purposes, especially for reaming and broaching, since individual lines may be cut out. This linearity can be partially suppressed by means of a rapid cooling during heat treatment, but it appears again when re-heating over the transformation point. The best possible surface quality is achieved furthermore by applying suitable cutting fluids, by changing the tool geometry (positive tool orthogonal rake angle) and through reducing the feed [Opit64, Peke74].

The applicable cutting conditions are little influenced by the heat treatment used – provided that machining is executed with cemented carbide tools and the tensile strength of the material is under 650 N/mm^2 [Well72]. Tools made of HSS, on the other hand, react more sensitively to differences in strength, so that different cutting speeds are required depending on the heat-treatment condition of the material.

Because of their very high toughness and low carbon content, case-hardened steels tend towards forming long chips. This lends especial importance to the selection of a suitable chip breaker when turning carbide indexable inserts. Chip breakage can be improved by alloying, for example, sulphur and lead (e.g. 16MnCrS5). Alloying sulphur reduces the strength of steel. This is because of the formation of manganese sulphides (see also Sect. 7.3.3). A reduction of strength by alloying lead is only observable from a temperature of over 300°C.

After machining, the case-hardening process follows in three steps: carburizing, hardening and tempering. The workpiece rim zones are carburized to obtain

0.6–0.9% carbon. After case-hardening (direct hardening, single hardening, double hardening), the hardness values in the rim zone increase up to 62 HRC. As a result of the warpage of the components caused by the case-hardening, a cutting post-machining process must be executed in some cases. Finest-grain carbides, mixed ceramics and PCBN cutting tool materials are especially suited to finishing highly heat-treated or hardened steels (> 45 HRC). Very high resultant forces arise in the process. Because the chip is annealed in the clearly high temperatures, chip breakage does not become a problem. As a rule, very high surface qualities are achieved in this way (see also Sect. 7.4.6).

Frequently used case-hardened steels include C15E, 16MnCr5, 20MoCr4, 18CrNi8.

7.4.3 Heat-Treatable Steels

Heat-treated steels have carbon contents between 0.2 and 0.6% and are therefore stronger than case-hardened steels. The main alloying components are silicon, manganese, chrome, molybdenum, nickel and vanadium.

The machinability of heat-treatable steels depends primarily on their crystalline structure, which is a result of the respective heat treatment applied, and can thus vary to a great extent. The influence of the material structure on machinability is generally stronger than that of the alloying elements.

The development of wear is essentially determined by the ferrite- and perlite-content in the microstructure. In the case of unalloyed heat-treatable steels with a carbon content up to about 0.5%, regularly formed ferritic/perlitic microstructures have a positive effect on machinability. Increasing the perlite content accelerates the development of tool wear and increases the resultant force. This limits the applicable range of cutting speeds. At higher cutting speeds, the end of tool life is brought about through crater lip breakage, especially when the steel contains larger amounts of chrome, manganese and vanadium (alloyed heat-treatable steels).

The length of the chips when machining heat-treatable steels depends considerably, as do other machinability criteria, on the respective type of heat treatment and the crystalline structure of the steel. Similarly to case-hardened steels, chip breakage can be improved by means of tool geometry and by alloying lead and sulphur. The intended use of the heat-treatable steels is of particular importance when selecting alloying elements which promote chip breakage, since this can lead to a reduction of strength.

Cementite spheroidization (soft annealing) is advantageous with respect to the wear behaviour of steels with larger ratios of carbon. This allows higher cutting speed to be used. However, the adhesion tendency also increases, which degrades the surface quality of the workpiece. Soft-annealed structures with a mixture of lamellar and granular cementite also lend themselves to machining at higher cutting speeds.

Abrasion and thermal wear become insignificant when machining heat-treatable structures (predominantly tempered martensite). The cutting speed should be reduced correspondingly. The most appropriate cutting edge materials for machining heat-treatable steels are coated cemented carbides for rough-machining and, for planing, cermets. Tools made of high speed steel are used in many cases for boring and thread die cutting. In view of abrasion and thermal wear, the selection of a suitable coating is recommendable here. Material hardness values > 45 HRC require highly wear-resistant cemented carbides, cutting ceramics or CBN as cutting materials.

For turning, milling and drilling, sulphur additives at a ratio of approximately 0.06-0.1% effect a clear improvement of the steel's machinability. At higher values, this improvement is reduced and the steel's strength is lowered.

Heat-treatable steels are heat treated in order to regulate their mechanical properties with respect to their intended purpose. They can only be adjusted for good machinability in few cases. Moreover, heat treatment procedures for achieving good machinability vary. For example, the machinability of steel C60E is improved through soft-annealing, that of steel C22E through coarse-grain annealing (or cold-forming).

In some cases, hardening and tempering takes place between rough-machining and planing or precision machining. Rough-machining, for which the most important factor is a high chip-removal rate, is performed on materials in a normalized state whose machinability is characterized, because of their ferritic/perlitic microstructure, by relatively low wear. Most components made of heat-treatable steel are machined in a heat-treated state. The associated strength values exceed those of the annealed state. For this reason, high cutting speeds induce a strongly increase in tool wear.

Among the heat-treatable steels frequently used in practice for machining are C45E, 42CrMo4, 30CrMoV9 and 36CrNiMo4. These materials are utilized for components of medium and high strain, especially in automobile and aircraft construction (connecting rods, axles, axle-pivots, rotor and crank shafts, springs, cogwheels).

7.4.4 Nitrided Steels

The carbon content of nitrided steels lies between 0.2 and 0.45%. They are heat-treatable and are alloyed with Cr and Mo for improved hardenability, as well as with aluminium or vanadium (nitride formers). Nitriding is carried out at temperatures between 500 and 600°C, i.e. below the α-γ-transformation temperature of the material [MDH80].

As opposed to case-hardened steel, for which high levels of hardness are achieved by means of a γ-α-phase transformation and the production of the metastable phase martensite, nitrided steel has a very hard surface traced back to the brittle metal nitrides. The nitrogen diffusing into the surface layer during the nitriding process forms with the alloying elements Cr, Mo and Al special nitrides. These mostly precipitate in submicroscopic form and cause high latticework tensions, i.e. high surface hardness.

The machining of these steels is done prior to nitriding, and usually in a heat-treated state. This structural state (i.e. fine, regularly distributed carbids, tempered martensite), which is favourable for subsequent nitriding, exhibits unfavourable machinability properties.

The heat-treated structure usually found when machining and the carbides distributed in the structure lead to high mechanical and thermal stresses on the tools. A short tool life is to be expected, especially at high cutting speeds. Due to the high strength values of heat-treated steels, the resultant force is relatively high.

If the nitrided steels are machined in a soft state, burr formation may cause a degradation of the surface quality and an impairment of the quality of the component.

When machining nitrided steels in a heat-treated state, one can expect predominately acceptable chip forms. In a non-heat-treated state, however, problems related to chip breakage will arise during machining.

Greater precipitations of ferrite in nitrided steel lead to an embrittlement of the rim zones and to an irregular transition in the core zone. Coarse-grain annealing for achieving good machinability is not to be recommended with respect to the steel's later use, since the ferrite would become even more coarse-grain and the strength would further decrease.

Nitrated steels with increased nickel content, such as 34CrAlNi7 with approx. 1% Ni, are difficult to machine. Nitrided steels containing aluminium are fundamentally more difficult to machine than aluminium-free ones, such as 31CrMo12, which exhibits a lower adhesion tendency. The addition of sulphur (34CrAlS5) has a positive effect on machinability. Nitrided steels are used in a similar range of applications as case-hardened steels (cogwheels, guide strips, etc.).

7.4.5 Tool Steels

One generally distinguishes between non-alloy and alloyed tool steels. Tool steels are required for different stresses. On this basis, the following categories are used:

- cold-working steels,
- hot-working steels and
- high-speed steels.

These categories are also useful for describing the machinability of tool steels.

Non-alloy tool steels in a forged or rolled state with a carbon content up to 0.9% contain lamellar perlite and ferrite, while those with higher carbon contents have lamellar perlite and a cementite network. Irrespective of carbon content, soft-annealed steels should have more or less regularly distributed cementite grains in a ferritic matrix. With an even higher carbon content, the cementite network cannot be removed by means of conventional soft annealing.

In a hardened state, the structure consists primarily of martensite in the rim layers. The martensite gradually transforms into intermediate structures as well as fine-lamellar perlite in the direction of the workpiece interior. In the case of supereutectoid steels, cementite grains are embedded in the matrix as well, if the steel was soft-annealed prior to hardening. Should this treatment be left out, then remnants of the brittle cementite network take the place of the cementite grains.

Non-alloy tool steels with a carbon content between 0.5 and 1.5% are machined in a soft-annealed state. Subeutectoid, non-alloy tool steels can also be machined in a normalized state or in the condition of delivery after hot working. In both cases, a relatively inferior machinability is to be expected because of the increased adhesion tendency and the growth of built-up edges.

The resultant force when machining tool steels is determined to a great extent by the special alloy composition and the type of heat treatment used. When machining alloyed tool steels, the dissolution of carbide formers and the increase in strength associated with this leads to an increase in the resultant force.

When machining tool steels in a normalized or soft-annealed state, the increased adhesion tendency and associated growth of built-up edges have a negative effect on surface quality. This can be partially remedied by means of a quenching and tempering to a higher strength. Because of the high deformability of ferrite, long chips with bad breakage form when machining tool steels in a soft-annealed state. An increasing carbide moulding degrades the chip breakage. If machining is executed in a heat-treated state, chip breakage is not to be considered a problem.

The amount of carbide formers is bears little importance for the machinability of alloyed tool steels. Carbide formers only increase the wear effect on the steel in an obvious way when they have dissolved during austenitizing and have not formed any carbides during subsequent annealing. The alloyed tool steels, especially high-alloyed high speed steels, are poorly machinable in an annealed state. This is due, as with unalloyed tool steels, to the marked formation of gluing-points and built-up edges. Disruptions may form at the outlet points of the tool. The adhesion tendency can be reduced by quenching and tempering to greater strengths (1200–1400 N/mm^2). This increases abrasive wear and the thermal stress on the cutting edge, however. The cutting speeds which are applicable in machining tool steels are, as a rule, relatively low and increase with the level of carbide moulding. However, the adhesion tendency of these steels, with finely distributed granular carbides, increases to an equal extent. Cutting materials most often used for machining tool steels are cemented carbides containing titanium carbides and tantalum carbides with medium toughness (e.g. from application group P20) as well as cermets. Following machining in an annealed state, tool steels can also be machined in a heat-treated state (R_m < 2000 N/mm^2) using cutting edge materials made of CBN.

Selecting alloying additives for tool steels is based first and foremost on their influence on surface hardness, hardness penetration depth, tempering consistency, toughness and wear resistance, whereby a suitable coordination with the carbon content is necessary, especially for higher-alloyed steels. The carbon content of the steel determines the ratio of carbides, which significantly promote abrasive wear. Carbon also influences hardenability and contributes decisively to tempering consistency and toughness via carbide reactions during hardening and tempering.

7.4.6 Hardened Steels

For a long time, grinding used to be the only way to machine hardened steels. The development of ultrahard cutting edge materials in the 1980's led to the machining of these materials with geometrically defined cutting edges. Since then, the machining of hardened materials exceeding a hardness of 50 HRC has come to be known as *hard machining*.

While hard turning had been implemented almost exclusively in single and small-batch production up to a few years ago, it is now used increasingly for large-batch and mass production. As a rule, hard turning is advantageous if a machining multiple sides can be achieved in a single clamping. When machining in one clamping, a very high accuracy of position of the functional surfaces can be realized. Hard turning procedures for machining interiors and contours are still very important.

For turning hardened steel, almost the only options are ultrahard non-metallic cutting edge materials, such as crystalline cubic boron nitride (PCBN) and – with some limitations – mixed ceramics. This is because of the extremely high requirements on hot hardness and the cutting tool material's consistency against diffusion resulting from the high temperatures and from cutting pressures in the machining zone. Because of their low hot hardness, however, finest-grain cemented carbides and ultrafine-grain cemented carbides are only suited to hard machining with low cutting times, such as turning with interrupted cutting, hard milling, broaching, gear skiving and -shaping. Because of their toughness, which clearly exceed both that of the very heat-resistant, though relatively brittle PCBN and, even more so, that of cutting edges made from mixed ceramics, they exhibit very good working behaviour in these applications [Koch96, Wina96].

The spectrum of chemical and physical properties of commercially available PCBN cutting materials is relatively large. In addition to clear differences in hardness and resistance to bending, properties that vary especially are heat conductivity, resistance to thermal shock and thermo-chemical inertia with respect to the material to be machined. This variance can be attributed to the cBN grain size and to the type and concentration of the binder phase [Kloc05b, Joch01].

When machining hardened steels with PCBN, the surface quality can be improved by modifying the cutting edge geometry. By selecting a large tool orthogonal clearance, one can principally increase the cutting time without changing the amount of free flank wear and simultaneously increase the displacement of the cutting edge. For the cutting edge fillet, an optimum radius of approximately 20 μm is recommended. Small bevels and radii lead to disruptions and wavering tool life, while larger bevels and radii cause larger forces and may induce oscillations [Kloc05c].

Mixed ceramics represent a further application group which can be used for hard machining with continuous cutting. In comparison to PCBN cutting edge materials, the use of mixed ceramics as cutting edge materials places less stress on the cutting edge with the result of clearly smaller chip thicknesses of a maximum 100 μm. The use of round indexable inserts or the selection of the greatest possible corner radius is recommended for pre-machining with smooth cutting [Abel93]. Because of the low toughness, anything but smooth cuts cannot be recommended [Köni89a, Momp87]. In consideration of the fact that tool costs are lower by a factor of 10–20, mixed ceramics for smooth-cutting represent a widespread alternative to cutting edge materials with a small ratio of CBN.

High resultant forces and high rake face temperatures arise during hard machining with rough cutting. Specific cutting forces k_c between 4000 and 4700 N/mm^2 and rake face temperatures which may even, in exceptional cases, exceed the local melting temperature of the material to be machined have been recorded [Berk92]. As opposed to soft machining, the passive force takes on very high values in this case. These values may even be higher than those of the cutting force [Köni84b, Töns93, Wobk93]. The normal compression stresses in the region of the wear mark on the side of the minor flank lead to a high mechanical and thermal stress on the workpiece rim zone.

The mechanical stress is the result of flank load similar to HERTZian pressure. The stress state in the workpiece caused by this pressure induces a residual austenite transformation and cold-forming in the rim layer of the workpiece [Gold91]. Through this process, compressive stresses are induced which increase with increasing flank wear of the minor cutting edge. For this reason, when hard-machining, the maximum amount of compressive residual stresses are shifted to the greater depths of the workpiece rim zone with an increase in flank wear.

The thermal stress is a result of the friction between the flank face and the workpiece. Both high normal stresses and high shear stresses lead, in combination with the relative movement between the tool and the workpiece, to high friction forces. High contact temperatures are the result, and these cause, via rapid cooling, the formation of martensite when the α-γ-transformation temperature is exceeded. The martensite is recognizable in the microsection as a "white", unetched layer (Fig. 7.17, right). A darker zone with a tempered structure can be seen below the white rim layer. Phenomena which occur at high temperatures because of transformations and high cooling gradients cause tensile residual stresses which are superposed on mechanically induced compressive residual stresses [Jonk87]. The magnitude of the flank face wear on the minor cutting edge has a significant influence on the distribution of stress in the workpiece rim zone. Metallographic studies

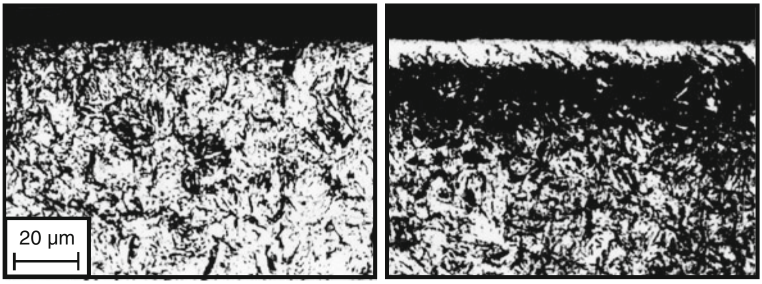


Fig. 7.17 Formation of the rim zone during hard turning

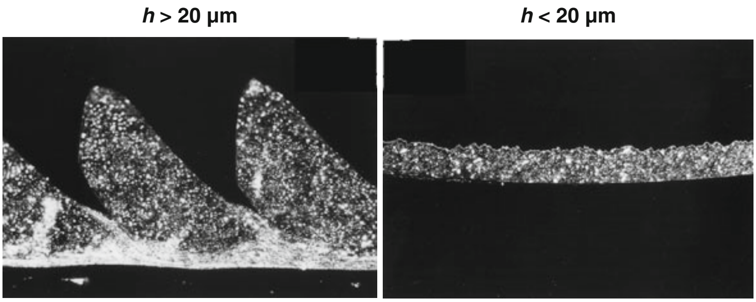


Fig. 7.18 Longitudinal section of a saw tooth chip (*left*) and a ribbon chip (*right*)

of hard-machined workpiece surfaces confirm the influence of tool wear on the formation of a new hardening zone. How the distribution of stress and the new hardening zones ultimately affect the behaviour of the component has not been sufficiently explained yet. Of particular interest are the effects on the fatigue strength of rolling contact stressed functional surfaces. The formation of new hardening zones during hard machining cannot be inhibited, not even by using coolants. Since the use of coolants effects no improvement of the tool life behaviour of the cutting edge material, hard-machining is generally performed in a dry state. When machining with interrupted cutting, the thermo-shock sensitivity of superhard cutting tool materials even forbids the use of coolants [Acke89, Töns91].

Highly heat-treated steels which receive their hardness mainly from their martensitic crystalline structure are practically non-deformable in room temperature and environmental pressure. This deformability behaviour results in a special chip formation, one different from the classic chip formation of unhardened steels. When machining hardened steels, two different chip forms may develop depending on the chip thickness. With high chip thicknesses ($h > 20\,\mu\text{m}$), a "saw-tooth" chip develops [Acke89, Berk92, Naru79] (Fig. 7.18), while smaller chip thicknesses ($h < 20\,\mu\text{m}$) tend to cause the formation of a ribbon chip [Köni93b].

The development of "saw-tooth" chips initiates material separation via a crack formation on the surface which transmits itself into the workpiece under an angle of the greatest shear stresses [Acke89]. The released chip segment slides off between the rake face and the newly developed separating surface. If the chip segment has slid so far that the increasing compressive stresses in turn cause the critical shear stress to be exceeded, the next crack forms. One theory ascribes the connection of chip segments to the fact that the crack is caught up in front of the cutting edge and a plastic deformation takes place due to the compressive stresses and high temperatures. Chip segments on a plasticized material thus slide off [Berk92]. In the case of high chip thicknesses, the transition from a ribbon chip to a saw-tooth chip is a function of the hardness of the material. With the predominately martensitically hardened, unalloyed tool steel C105W2 (1.1645), for example, this transition takes place at approximately 50 HRC [Naru79]. At low chip thicknesses ($h < 20\,\mu\text{m}$), the chip thickness is in the region of the cutting edge radius r_n with a tool orthogonal rake angle which acts in a correspondingly strongly negative way. A triaxial stress state with a high ratio of hydrostatic compressive stress is thereby caused in the area of the material directly in front of the cutting edge with no material separation ensuing. The shear stress hypothesis put forward by MOHR offers an explanation of these phenomena. VON KÁRMÁN already indicates the possibility of a plastic deformation of brittle materials under pressure on all sides [Karm11]. If such a state of compressive stress exists, the shear yield stress becomes a decisive criterion. If, however, the strain on the material is characterized by a monoaxial compressive stress state as it occurs on the workpiece surface, the failure of the material is caused by a cleavage fracture.

Precision hard turning ($v_c \approx$ 100–200 m/min, $f \approx$ 0.05–0.15 mm, $a_p \approx$ 0.1–0.5 mm) is used for manufacturing ready-to-use components. Surface roughness values in the region of R_z= 2.5–4 μm can be achieved in series manufacturing in a procedurally reliable way. A further development of precision hard turning tending toward higher component quality is *high-precision hard turning* ($v_c \approx$ 150–220 m/min, $f \approx$ 0.01–0.1 mm, $a_p \approx$ 0.02–0.3 mm), which has the potential of achieving surface roughness values which lie clearly below those of conventional precision hard turning processes. The surface roughness R_z achievable when using special machine tools lie below 1 μm. However, one must also expect that increasing tool wear will bring about a degradation of surface quality. The procedure shows potential for creating surfaces with $R_z \leq 3$ μm with long tool life.

7.4.7 Non-rusting Steels

Non-rusting steels contain at least 10.5% chromium and a maximum of 1.2% carbon. Their main alloying elements are chrome and nickel. Amounts of chromium exceeding 12% render the steel material corrosion-resistant. Nickel expands the γ-region and, in the case of steels with a high chromium content, leads to the stabilization of the austenitic structure, which is generally unstable in the case of low-alloyed carbon steels and which disintegrates into ferrite and cementite below the Ac1-temperature. Nickel starkly decreases the heat conductivity of the steel material. "Chrome steel" is the usual term for ferritic steel types and "Cr-Ni steel" for austenitic steel types. According to their essential usage properties, DIN EN 10088-1 categorizes non-rusting steels into:

- corrosion-resistant steels,
- heat-resistant steels and
- high-temperature steels.

Non-rusting steels are distinguished by good persistence against chemically aggressive substances. In general, they have a chromium content of at least 12%. Protection is achieved with a chromium content of more than 10.5% by means of the spontaneous formation of a protective chromium oxide layer [DINEN10088]. With respect to their metallographic constituents, the corrosion-resistant steels are subdivided into ferritic, martensitic, austenitic and ferritic-austenitic steels. Figure 7.19

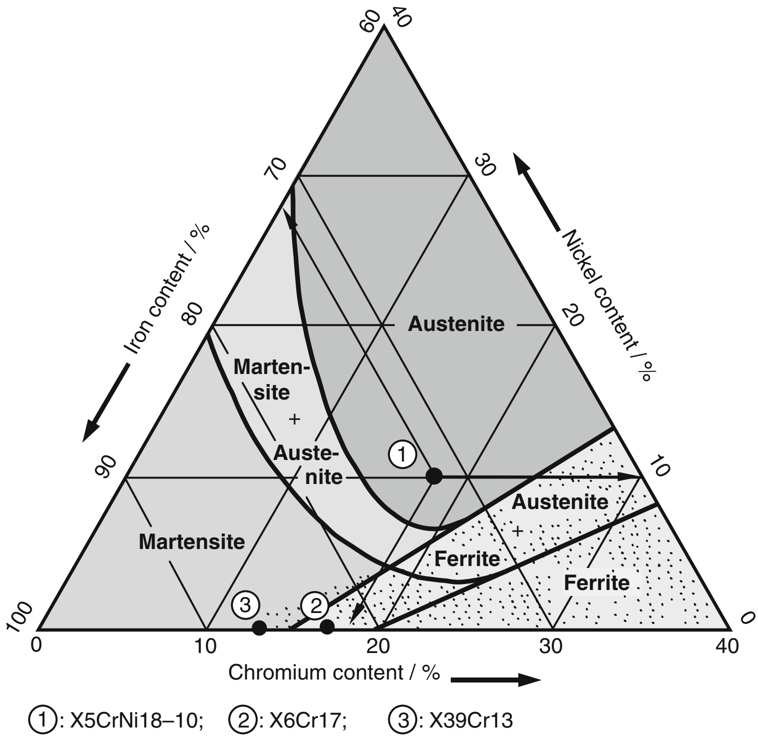


Fig. 7.19 Phase diagram of Cr-Ni-steels with metallographic constituent data after quenching, acc. to IGNATOWITZ [Igna97]

shows the phase diagram of Cr-Ni steels. One can see from this diagram which structure is to be expected, depending on chromium and nickel content, with a non-rusting chromium-nickel steel after quenching. Point 1 corresponds to the austenitic steel X5CrNi18-10 with 18% chromium, 10% nickel and 72% iron, point 2 corresponds to the ferritic chrome steel X6Cr17 and point 3 to the martensitic steel material X39Cr13 [Igna97].

The ferritic steels are primarily chrome steels with a chromium content of 12.5–18% and a C-content below 0.1% (e.g. X6Cr13, material no.: 1.4000, X6Cr17, material no.: 1.4016). They are magnetic and not hardenable.

Martensitic steels are primarily chrome steels with a chromium content of 12–18% and a C-content of 0.1–1.2% (e.g. X12Cr13, material no.: 1.4006; X39Cr13, material no.: 1.4031). Depending on the quality, these steels also contain additional Ni and Mo. They are magnetic and can be heat-treated and tempered by means of a corresponding heat-treatment. With increasing C-content, the hardness of the steel increases when in a hardened state. A hardness of 40 HRC is achievable with a C-content of 0.1% and a hardness of 59 HRC with a C-content of 0.9%.

The steels the most used by far are austenitic steel materials. They contain approximately 17–26% Cr, 7–26% Ni, less than 0.12% C and, in some cases, small amounts of Si, Mo, V, Nb, Ti, Al or Co (e.g. X5CrNi18-10, material no.: 1.4301; X6CrNiMoTi17-12-2, material no.: 1.4571). They are not magnetic and cannot be hardened by means of heat treatment. In an annealed state, they are characterized by very good toughness properties which are retained even at extremely low temperatures. Especially given a high carbon content, they tend when cold-formed towards considerable strain hardening. Through cold-forming, therefore, their strength values can be increased dramatically. When strain-induced martensite forms, however, the strain is reduced to a very considerable degree.

Among the products made from the abovementioned corrosion-resistant steels are household devices, razor blades, knives, surgical instruments, parts for automobile construction, agricultural and materials handling technology, mechanical and plant engineering and devices and instruments for the food industry, the chemical industry, the textile industry and for shipbuilding [Stah01].

A further group of corrosion-resistant steels are the ferritic-austenitic steel materials, also either referred to as duplex steels (e.g. X2CrNiMoN22-5-3, material no.: 1.4462) or super duplex steels (e.g. X2CrNiMoCuWN25-7-4, material no.: 1.4501). The name "duplex" refers to their two-phase crystalline structure consisting of ferrite and austenite. These steels have optimal properties at a balanced ferrite/austenite ratio of approximately 50/50%. In comparison to normal austenitic steels, the duplex steels contain less nickel (about 4–8%), though usually a significantly higher amount of chromium (about 18–25%). In order to increase their resistance to intercrystalline corrosion, a certain amount of nickel is exchanged for additional nitrogen as an austenite former and molybdenum. The optimal microstructure is created by means of a heat-treatment at 1000–1100°C. These steels find the most use in the gas and oil industry, in the petrochemical industry, in chemical tankers and in sewage treatment plant construction. In comparison to austenitic rustproof steels, the duplex materials have a strain limit which is almost double as high ($R_{p0,2}$ about 400–550 MPa) given similar or markedly higher strength values.

Heat-resistant steels are mainly ferritic and austenitic steels with a high resistance against oxidation as well as against the influence of hot gases and combustion products above 550°C. As a rule, the heat-resistant ferritic steels contain at least 12% chromium as well as aluminium and silicon (e.g. X10CrAlSi13, material no.: 1.4724; X10CrAlSi25, material no.: 1.4762). Heat-resistant austenitic steels are additionally alloyed with at least 9% nickel (e.g. X8CrNiTi18-10, material no.: 1.4878; X15CrNiSi25-21, material no.: 1.4841). These steel types are used, for example, in ovens and apparatus construction for annealing bells, baskets and tubes [Stah01].

The high-temperature steels are mainly martensitic and austenitic steel types with a high long-time rupture strength with a long-term mechanical strain above 500°C. Among the products made from high-temperature martensitic steels (e.g. X20CrMoV11-1, material no.: 1.4922; X20CrMoWV12-1, material no.: 1.4935) are components for thermal power plants, steam boilers and turbines, as well as for the chemical industry and nuclear technology. High-temperature austenitic steels (e.g. X6CrNi18-10, material no.: 1.4948; X5NiCrTi26-15, material no. 1.4980) are used in the construction of pressure vessels and armatures, pressure tanks and steam boilers, but also in the manufacture of blades, discs, axles and bolts for steam and gas turbines as well as in shipbuilding [Stah01].

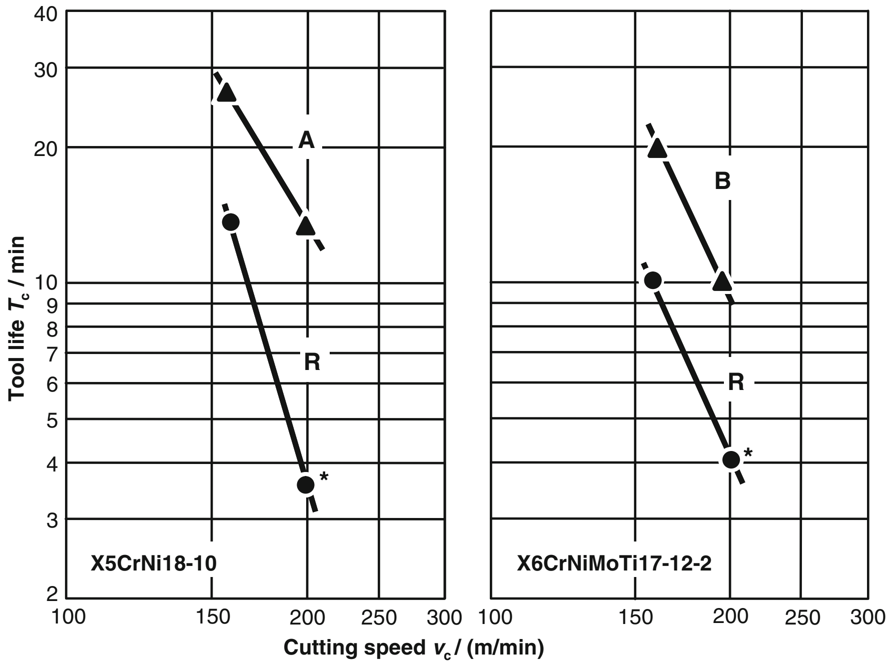
Non-rusting steels with a ferritic structure lend themselves to machining relatively well. The onset of wear through abrasion and adhesion is comparatively low. The machinability of martensitic steels is a function of hardness, which is in turn a result of the heat treatment used. Depending on the heat treatment, the structure consists either of martensite (hardened) or tempered martensite with chromium carbides and ferrite (quenched and tempered). Duplex steels are considered to be extremely hard to machine. Characteristic reasons for this are a marked adhesion tendency, high strain hardening of the workpiece rim zone and an unfavourable chip formation.

Because of the eminent importance of corrosion-resistant austenitic steel materials, the following will go into more detail on their machinability properties. Austenitic steel materials are machined in either a quenched or a solution-annealed state. In comparison to ferritic-perlitic or quenched-and-tempered steels, they are significantly harder to machine. This is due to their high deformability and toughness, their tendency towards strain hardening and towards adhesion with the cutting material, as well as their lower heat conductivity, which is about 1/3 lower than with non-alloy steels. The latter impairs heat removal through the chip and increases the thermal stress of the cutting edge. In spite of the comparatively low tensile strength of austenitic Cr-Ni steels, the results of these specific material properties are a high thermal stress of the tool cutting edge, pronounced flank face and/or rake face wear, material gluing, notch wear, pitting of the cutting tool material, cutting edge breakages and unfavourable chip forms, which allow for machining tasks to be performed only at relatively low cutting speeds, for short tool lives and with insufficient surface qualities. In the case of coated tools, one can observe in many cases that the adhesion between the material and the hard material coating is stronger than their adhesion on the substrate and that this leads to the delamination of the coating.

Austenitic steel materials are machined preferably with uncoated or coated WC-Co cemented carbides of the main application group M. Because of the high thermal stress of the cutting edge, cutting speeds are used which, in comparison to those used in turning ferritic-perlitic steels, are lower by a factor of 2–5. Applicable cutting speeds range from $v_c = 50$ m/min (X5NiCrTi26-15) to $v_c = 160$ m/min (X6CrNiMoTi17-12-2), depending on the alloy. The tool life is between 5 and 15 min.

A type of wear which limits tool life when turning austenitic steel materials with uncoated cemented carbides is the formation of a pronounced crater on the rake face (Fig. 3.41, left). Due to this strong crater wear, uncoated cemented carbides can only be used at relatively low cutting speeds ($v_c < 100$ m/min) [Gers04]. The use of high cutting speed is only made possible when turning austenitic steels through the application of coated tools (Fig. 3.41, right).

Significantly improved performance in turning austenitic steels requires cutting tool materials with properties customized with respect to the respective machining task (Fig. 7.20). The more the cutting tool material system reaches the physical limits of the substrate and the hard-material coating in the chipping process, the



Cutting tool materials:

A: WC-7.5Co-6.8MK-0.75Cr_3C_2 / TiN-TiCN-Al_2O_3-HfCN

B: WC-7.5Co-6.8MK-0.75Cr_3C_2 / TiAlN

R: Reference cutting tool material

Tool life criterion:

flank wear VB = 0.3 mm, *: plastic deformation of cutting edge

Process: external cylindrical turning

Cutting parameters: feed: $f = 0.32$ mm, depth of cut: $a_p = 2$ mm

Cutting fluid: emulsion 6%

Fig. 7.20 Examples for the performance of carbides with tailored substrates and coating systems compared to a common cutting tool material during roughing of austenitic steel, acc. to GERSCHWILER [Gers04]

more carefully the substrate, the hard-material coating and the tool geometry must be customized to the individual machining task [Gers04].

In addition to a high resistance to wear, machining tools used to turn austenitic steel materials are also required to exhibit “performance” in terms of chip formation and chip breakage. Because of their high deformability, the austenitic steels tend more than any others towards the formation of long ribbon and snarled chips. The latter represent a high hazard potential not only for the component and the machine, but also for the operating personnel. It is thus essential, and not only in automated manufacturing plants, that this type of chip formation is avoided through all available means to secure a higher level of machining safety.

A safe chip breakage is guaranteed in two ways. Firstly, on the material side, through alloying measures. In many ways, however, there are narrow limits on these measures placed by the usage properties of the material and the component. The more promising way in most cases is thus the optimization of the cutting part geometry. In case of varying cutting depths and/or small feeds, however, this places an extremely high demand on the geometric design of the cutting edge. Especially with small chip cross-sectional areas, the tools must possess chip shaping components which lie close enough to the cutting edge so that they come into any contact at all with the fine chips [Köni90a].

The high ductility of austenitic materials requires tools with the sharpest possible cutting edges. Sharp tool cutting edges facilitate the separation of the material during chip formation and, by reducing the resultant forces, contribute in a highly essential way to lowering the plastic deformation of the workpiece rim zone, to the creation of high-quality workpiece surfaces and to the reduction of burr formation. On the other hand, however, one can also observe that, with increasing mechanical stress, very sharp cutting edges lead also contribute to an increased pitting of the cutting tool material, which in turn leads to accelerated wear. As experiments have shown, a defined cutting edge fillet can stabilize the cutting edge and significantly improve tool life. When finish turning with small chip cross-sectional areas, however, a cutting edge fillet no greater than 30 μm should be selected. When rough turning, on the other hand, a highly stable cutting edge can be designed with a 40–60 μm fillet in order to prevent edge breakages and pittings of the cutting tool material.

Burr formations become a further serious problem when machining ductile austenitic steel materials. In the case of components which are to be burr-free, up to 20% of manufacturing costs may result from burr removal [Köni93]. In addition to the costs for burr removal, increasing demands on component quality and workplace attractiveness entail combating burr formation through suitable measures.

As a rule, austenitic steels are machined by wet cutting. However, dry machining is thoroughly possible with corresponding process dimensioning. One must consider, however, that due to the lack of the cooling effect of the cutting fluid, both chip and tool exhibit higher temperatures than is the case with wet machining. The results of this are an increased in smeared material on the tool, clearly inferior chip formation, more aggressive notch wear on the major and minor cutting edges and, as a result, increased burr formation and inferior surface qualities. In comparison to wet machining, dry machining austenitic steels can lead to a drastic reduction of tool life, the latter being limited as a rule not by flank face wear, but rather by the degradation of the surface quality associated with notch formation on the minor cutting edge. Machining austenitic steels in an economical way thus requires special measures. In addition to a tool geometry customized to the machining task, suitable coating systems and cutting parameters, the most important of these is the use of a minimum quantity lubrication. This lubrication in particular can reduce rewelding of the material and significantly improve the surface quality. Mediums used for this are, as a rule, synthetic ester oils [Gers04] (also Chap. 6).

7.5 Machinability of Cast Iron

Cast iron refers to iron-carbon alloys which have a carbon content above 2.06%. In the iron-carbon two-material system, cast iron becomes a steel (up to 2.06% C). The material is usually shaped by means of casting and a final machining operation.

Cast-iron alloy material groups are classified on the basis of the appearance of their fracture faces into white and grey cast iron (Fig. 7.21). Among the white cast iron materials are chilled cast iron and malleable cast iron, both of which solidify according to the metastable iron-iron carbide system and in which carbon is thus present in the form of cementite (Fe_3C). As opposed to this, the solidification of grey cast iron takes place according to the stable iron-graphite system, with the carbon thus being present as graphite. Depending on the form of the graphite, grey cast iron materials are divided into cast irons with lamellar graphite (GJL), cast irons with vermicular graphite (GJV) and cast irons with spheroidal graphite (GJS).

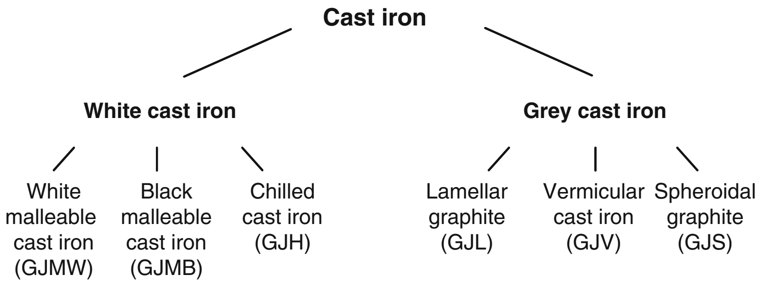


Fig. 7.21 Cast iron classification

In addition to the metallic matrix, the machining properties of iron-cast materials are also heavily influenced by the amount and formation of the embedded graphite. Graphite inclusions, in the first place, reduce friction between the tool and the workpiece and, on the other, disrupt the metallic matrix. This leads to improved machinability in comparison with graphite-free cast iron or steel materials. The results are short-breaking chips, small grinding forces and higher tool lives.

The metallic matrix of cast iron is, to a high degree, crucial to its machinability. The matrix is influenced by the respective chemical composition (alloying elements) and heat treatment and consists, in the case of materials with low strength, predominately of ferrite. The material becomes stronger as the amount of perlite rises, which also leads to a more significant abrasive tool wear. Cast irons of high strength and hardness often have a bainitic, ledeburitic or martensitic structure and are therefore very hard to machine.

7.5.1 White Cast Iron

As mentioned above, the solidification of white cast iron takes place according to the metastable Fe-Fe_3C system (Fig 7.22). If the cooling speed is high, a eutectic

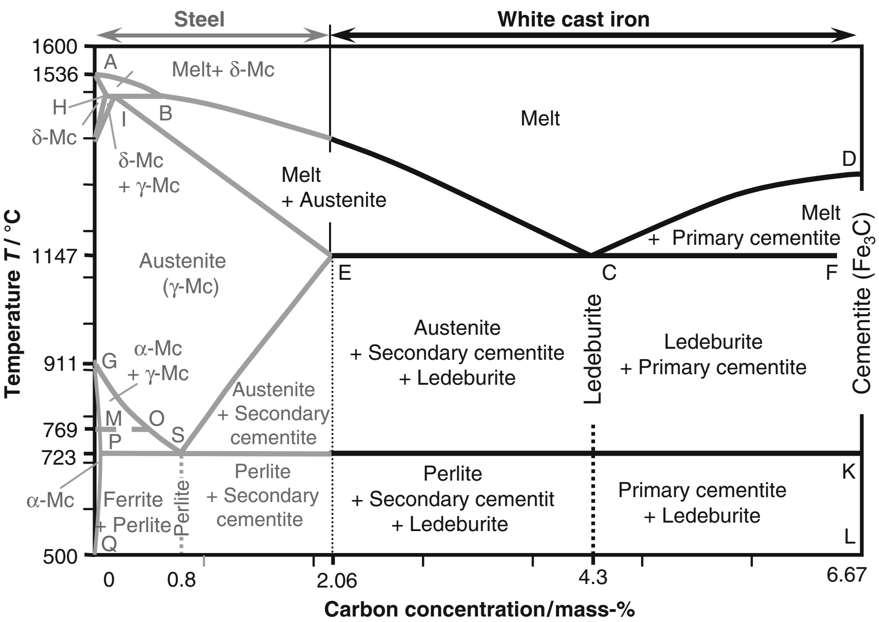


Fig. 7.22 Section of the iron-carbon phase diagram (metastable system, fast cooling)

(1147°C, C-content = 4.3%) forms which is referred to as ledeburite (austenite + cementite). Depending on the carbon content, additional perlite and secondary cementite is found in hypoeutectic compositions and additional primary cementite in hypereutectic alloys.

The high ratio of iron carbide in white cast iron leads to great hardness with a pronounced level of brittleness. If the white cast iron is not subjected to further heat treatment, it is referred to as chilled cast iron. If a further heat treatment is executed (tempering), malleable cast iron forms.

7.5.1.1 Malleable Cast Iron

The usage properties of malleable cast iron are adjusted by means of a metastable solidification and subsequent heat treatment (tempering). On the basis of the fracture appearance, malleable cast irons are subdivided into white malleable cast irons (GJMW) and black malleable cast irons (GJMB).

In order to produce white malleable cast iron, the white cast iron is annealed in an oxidizing atmosphere (approx. 1000°C, 60–120 h). The aim is to produce a partially or completely decarburized material by removing the carbon from the casting [DINEN1562]. The degree of decarburization of the material is strongly dependent on the duration of the tempering process and on the section thickness of the casting. A complete, uniform decarburization is only achievable with low section thicknesses; with thicker castings, a rim decarburization and a disintegration of the cementite in the workpiece interior take place. The remaining graphite is then present as temper carbon. In the case of partial carburization, therefore, there is a purely ferritic rim zone structure and a perlitic-ferritic core structure.

As opposed to white malleable cast iron, black malleable cast iron is heat-treated in an oxygen-free atmosphere. The cast iron is thus not decarburized. The cementite disintegrates entirely into temper carbon [DINEN1562]. A characteristic property of black malleable cast iron is that, because of the non-decarburizing annealing, there is a uniform structure across the workpiece cross section, irrespective of the section thickness. As a function of cooling speed, a ferritic matrix forms given slow cooling, with the temper carbon evenly distributed in the form of nodular formations. Given a more rapid cooling, a perlitic or even martensitic matrix forms. Perlitic-ferritic mixed matrices are also possible.

In contrast to steels, malleable cast iron is excellent for machining purposes. In spite of the good plastic deformability of the different types of malleable cast iron, the manganese sulphides and temper carbons embedded in the steel-like ferrite or perlite matrix cause short-breaking chips which can be removed without difficulty [Wern83]. Moreover, the temper carbon reduces the friction between the workpiece and the tool and interrupts the base metallic material, which leads to low resultant forces and higher tool life. Given identical workpiece hardness, black malleable cast iron is more easily machined that white. This circumstance is attributable to white malleable cast iron's purely ferritic rim zone structure (without temper carbon). Problems arising when machining ferrite are the growth of built-up edges (adhesion tendency of ferrite), the formation of ribbon and snarled chips (high deformability of ferrite) and the reduction of the surface quality with increased burr formation.

Cutting tool materials usually employed for machining malleable cast iron are uncoated and coated cemented carbides, cermets, oxide ceramics and PCBN of the main application groups P and K.

As a result of heat treatment, a remarkable consistency is achieved in the production of malleable cast irons with respect to their mechanical/technological material properties, which allows optimal cutting conditions for economic manufacturing. Furthermore, malleable cast irons are characterized by high strength and toughness properties as well as excellent casting processability. Workpieces can be achieve with accuracy of form with high surface quality, even thin-walled and intricate workpieces. Because of their high ductility, malleable cast irons are especially suitable for components which are subjected to dynamic stresses – oscillating or impulsive – and which must resist mechanical forces of great magnitude [Wern00].

White malleable cast iron is both weldable and hardenable. When selecting a hardening procedure, one must consider that white malleable cast iron exhibits a structure which is dependent on section thickness and a carbon content which increases from the outside to the inside. Thus only a thermochemical procedure can be applied for the low-carbon rim layer, e.g. case-hardening or nitriding [Schü06].

7.5.1.2 Chilled Cast Iron

White malleable cast iron without heat treatment is referred as a result of its high zementite-content as chilled cast iron (GJH). It is both very hard and brittle and thus suitable for components which are to exhibit a high resistance to wear with low dynamic stresses, e.g. for rollers and grinding tools.

Because the cementite is very hard, chilled cast iron is very difficult to machine. Tool cutting edges are subjected during machining to high mechanical and thermal stresses. Therefore, high demands must be made on the cutting tool material with respect to wear resistance and pressure resistance. Materials used for machining chilled cast iron are, almost exclusively, cemented carbides, and, especially with high hardness values, cutting ceramics (oxidic mixed ceramics) and PCBN cutting tool materials of the main application group H. In comparison to cemented carbides, the use of cutting ceramics allows for an increase of the cutting speed by a factor of 3–4. The higher material removal rate, however, comes with the disadvantage of an increased susceptibility to breaking.

In order to keep the mechanical stress of the tool cutting edge as low as possible when machining chilled cast iron and thus to improve the tool life parameters, the cutting speed and the feed should be reduced with increasing material hardness. Commonly used in chilled cast iron machining are a lead angle of 10–20° and a tool orthogonal rake angle of –5 to 5°.

7.5.2 Grey Cast Iron

In grey cast irons, the carbon is ideally present in the structure exclusively in the form of graphite. The melt solidifies according to the stable iron-graphite system (Fig. 7.23). In reality, however, because of the final cooling speed, small amounts

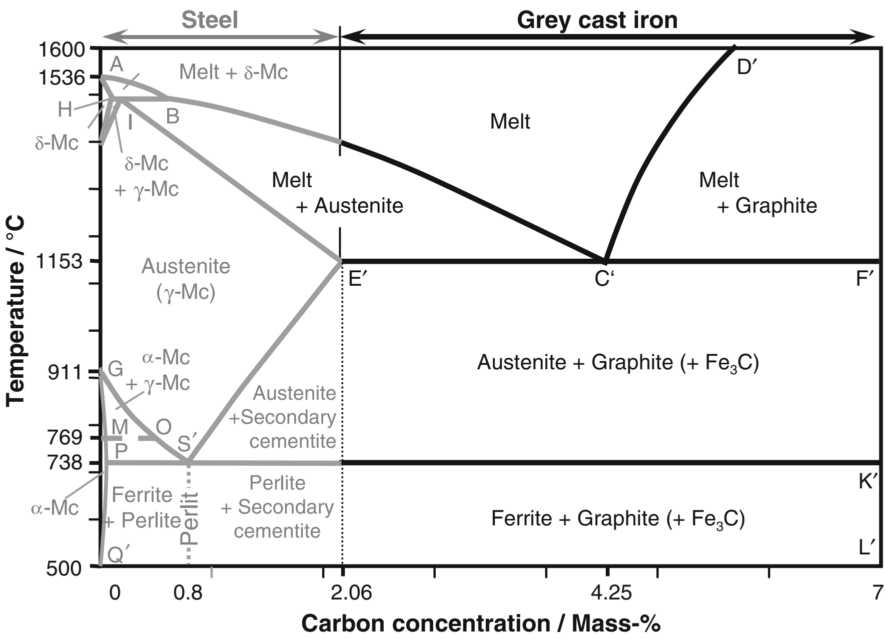


Fig. 7.23 Section of the iron-carbon phase diagram (stable system, slow cooling)

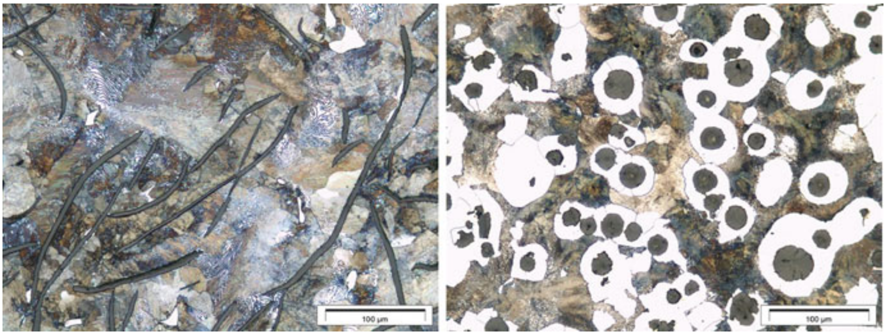


Fig. 7.24 Cast iron with lamellar graphite (*left*) and spheroidal graphite (*right*)

of Fe_3C are found in the structure in addition to graphite. In order to maximize the amount of graphite, so-called graphitizing elements such as silicon can be added to the melt.

As mentioned above, both the mechanical properties and the machinability properties of cast iron materials containing graphite depend in a very clear way on the crystalline structure of the matrix and the form of the graphite inclusions.

The type of metallic matrix can vary broadly according to the selection of chemical composition and heat treatment. Cast iron types with a ferrite matrix have the lowest strength and the highest plasticity and toughness; the types with a perlitic matrix have the highest strength and the lowest plasticity and toughness. Types with a mixed matrix (with different ferrit-perlite percentages) fall somewhere between these limits [Herf07]. In practice, the matrix is usually either ferritic-perlitic or purely perlitic. Only after an especially long annealing process does a purely ferritic matrix form.

The graphite inclusions take on varying forms depending on the conditions of their origination, which are also controllable through metallurgical measures (Mg-content). These forms vary from disc-like forms (lamellar) to fissured forms (worm-like, i.e. vermicular) to globular particles (Fig. 7.24) [Hors85].

One can say in sum that the machinability of grey cast irons is not determined by the hardness and strength of the alloy alone; the form, quantity and distribution of the graphite inclusions in combination with the given type of matrix play an important role [Stäh65, Töns91]. The following will treat the different types of grey cast iron.

7.5.2.1 Cast Iron with Lamellar Graphite

Cast iron with lamellar graphite (GJL) is the most frequently produced type of cast iron. The graphite is present in the form of thin, unevenly formed discs referred to as “graphite lamellae”. These graphite lamellae interrupt the base metallic mass to a very great extent. Also, when stress caused by external forces is present, stress peaks form at the margins of these graphite lamellae. The lamellae then act as internal notches (predetermined breaking points), for which reason the tensile strength of this type of cast iron is relatively small (100–350 N/mm^2), the material being designated as brittle [Herf07a]. On the other hand, lamellar graphite gives the material a high heat conductivity, favourable damping properties and, due to the embrittlement, high stiffness. Cast irons with lamellar graphite are especially suitable for machine beds and stands and bearing, gear and machine housings.

When machined, GJL is distinguished by excellent self-lubricating properties. These result from the circumstance that the graphite lamellae are cut during the machining process, with the graphite consequently forming a lubrication layer on the tool. This leads to lower wear and higher tool life. In addition, the graphite lamellae act favourable on the chip form, since they stop any incipient shear early on and induce cracking leading to the formation of segmented or discontinuous chips. Short-breaking chips develop, usually spiral chip segments or discontinuous chips.

The mechanical strain on the tool cutting edge when machining GJL tends to be low, leading to low resultant forces. A major advantage when machining GJL is derived from the formation of wear-reducing manganese sulphide layers on the flank and rake faces of the cutting insert. With increasing cutting speeds, the manganese sulphides form a layer on the insert which becomes thicker and thicker. This layer reduces the friction coefficient on the one hand and acts as a diffusion barrier and a protection from wear on the other. The formation of wear-reducing coatings only sets in with higher cutting speeds (from approximately 200 m/min onwards), i.e. due to the resultant increase in the machining temperature. This effect ensures long tool lives when machining GJL.

The surface quality of machined workpieces made of GJL is a function of the finishing method, the cutting conditions and the composition, fineness and evenness of the cast structure [Opit70]. There is generally no burr formation at the workpiece edges during machining; instead, because of the material brittleness, there are edge disruptions.

The hardness of the material is a reference value in the first approximation for the applicable cutting speed. Cast iron with lamellar graphite only having a small amount of perlite (about 10%) after annealing treatment, for example, can be machined, with the same tool life, at a cutting speed three times faster than with a cast iron with a large amount of perlite (about 90%). Other hard metallographic constituents, e.g. the phosphide eutectic steadite, increase tool wear and reduce tool life in the same way cementite does in perlite. They significantly reduce the applicable cutting speeds (Fig. 7.25).

The rim zone structure of cast iron workpiece generally exhibits lower machinability than the core zone. This can be attributed, on the one hand, to non-metallic inclusions and, on the other, to the altered graphite structure and microstructure directly beneath the outermost cast layer, as well as to high temperature oxidations. This results in stronger abrasive wear and the formation of a wear notch on the tool cutting edge. In practice, this is often compensated by means of a reduction of the cutting parameters.

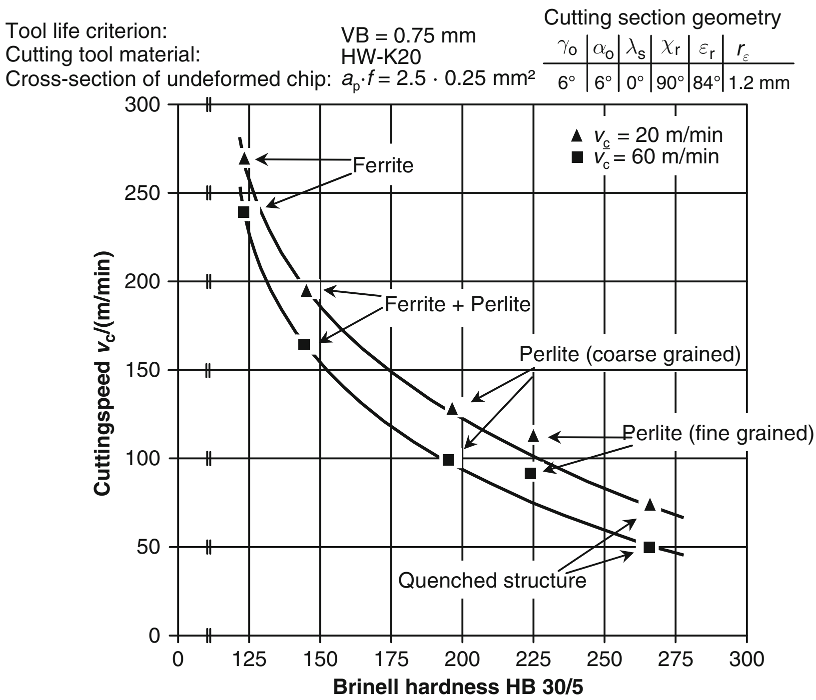


Fig. 7.25 Influence of the crystalline structure and hardness on the cutting speed during turning of cast iron with lamellar graphite

A rich array of cutting tool materials lends itself to machining cast iron with lamellar graphite. The use of high speed steels is generally limited to tools with very fine cutting edges. Typical procedures used are boring, reaming and thread die cutting. The classic types of cemented carbide for machining cast irons with lamellar graphite are those in the main application group K. Types K01-K05 and especially cermets are especially suited to precision and ultra-precision machining. The material removal rate can be increased considerably by coated cemented carbide, cutting ceramics and especially PCBN cutting tool materials of the main application group K.

Coated cemented carbide indexable inserts are used for turning, boring and milling grey casts. The longest tool lives and highest cutting speeds can be achieved in the process by using cemented carbide with ceramic multi-layer coatings. Silicon nitride ceramics are suitable for rough machining with large interrupted cuts or extreme irregularities in the contour of the workpiece as well as for machining with cutting fluids.

7.5.2.2 Cast Iron with Vermicular Graphite

In cast irons with vermicular graphite (GJV), the graphite doesn't form in lamellar or globular form, but rather in coral-like form, worm-shaped in microsection (lat. Vermicular: worm-shaped). The rounded edges of the graphite inclusions are reduced by internal stress peaks, as they appear on the tapered ends of graphite lamellae in GJL.

With respect to its mechanical and thermal properties and its machinability, vermicular cast iron lies between cast iron with lamellar (GJL) and with globular (GJS) graphite inclusions. In comparison to lamellar cast iron, GJV has higher strength (up to 70% higher), higher toughness, higher stiffness, higher hardness, higher endurance strength, higher oxidation-resistance and higher thermal shock resistance. In comparison to cast iron with spheroidal graphite, GJV has better pouring properties, a better machinability, a better damping ability, a lower tendency to warp, a lower thermal expansion and a better deformation resistance with temperature change. Because of these properties, cast irons with vermicular graphite can be cast more with thinner walls than GJL and GJS, which allows for lighter components. These combinations of properties render GJV suitable as a construction material for combustion engines. These highly resistant types of cast iron are particularly suited to the demands of modern direct injection diesel engines. Particularly when faced with the conflicting interests of high component strength vs. low component weight, GJV offers a good alternative to traditional construction materials such as aluminium and cast iron with lamellar graphite. In comparison to traditional motors made with GJL, the use of GJV allows for a weight reduction of up to 20%. In addition to vehicle-related areas of application, such as engine blocks, crankcases, cylinder heads, cylinder liners, exhaust manifolds, brake and clutch discs, there is a growing general interest in using GJV in other applications than those in automobile construction [Kloc01, Leng06, Röhr06].

The industrial implementation of GJV in the automobile industry is slow because of the uneconomical and difficult machinability of this material in comparison with cast iron with lamellar graphite. The main difference between the machinability of GJL on the one hand and that of GJV and GJS on the other is that, when machining GJL, wear-reducing manganese sulphide coatings form on the flank and rake faces of the cutting insert. When machining cast iron with vermicular and spheroidal graphite, no sulphur and no manganese sulphide is present and thus no wear-reducing layer. This is because of the magnesium treatment necessary for the formation of graphite. With lower cutting speeds (v_c < 200 m/min), at which, because of the low temperatures, no wear-protective coating develops when machining GJL, tool life travel path differences when machining different cast iron materials are essentially determined through differences in the mechanical characteristic values (hardness, tensile strength) as well as in the graphite morphology.

An essential impact on tool life is made by alloying elements, e.g. titanium. The alloying element titanium forms titanium carbides which are harder than WC-Co cemented carbides and thus cause an increase in the abrasive wear of the tool. Studies have shown that doubling the titanium content reduces the tool life travel path by about half. In GJV types used most frequently at present, the titanium content is less than 0.015% [Kopp04].

When machining GJV, chip formation is discontinuous within a broad range of cutting speed, with the result of short-breaking chip forms. The inhomogeneity of the material resulting from the different mechanical properties of the phases ferrite, perlite and precipitated graphite encourage this form of chip development [Röhr06].

Among the cutting tool materials for machining cast iron with vermicular graphite, coated cemented carbides (HC) and, in particular, aluminium oxide ceramics (CA) show great potential for machining GJV. When comparing the tool life travel paths of HC and CA at different cutting speeds, however, one can see that each of the two cutting tool materials is suited to a respective cutting speed range. Coated cemented carbide is suitable for machining at conventional cutting speeds, while aluminium oxide ceramics is the preferable cutting tool material for high-speed machining. Because of the dynamic excitation caused by the discontinuous chip formation, tougher WC-Co-based cutting tool materials can better compensate for the high stress reversals in lower cutting speed ranges than ceramic cutting tool materials. At higher temperatures, however, the hot hardness and the chemical resistance of the cutting tool material become decisive factors, giving CA tools good usage behaviour at high cutting speeds and temperatures at which HC tools become weak and thus fall victim to increased chemical and abrasive wear [Röhr06].

Generally, because of their high hardness and wear-resistance, PCBN cutting tool materials allow for significantly cutting speeds and a correspondingly higher productivity compared to cemented carbides. Even when using PCBN as cutting tool material, because of the absence of an MnS protective layer when machining GJV, the optimal cutting speeds for machining GJV lie at an approximate value of 300 m/min, clearly lower than those used for GJL (up to 1500 m/min). If high cutting speeds are used (500–1000 m/min), significantly reduced tool live travel paths can be expected compared to GJL [Leng06, Reut02].

Initial studies have shown that PKD appears to be an interesting cutting tool material, one with which the tool lives achieved when machining GJV are comparable to those for GJL. In comparison to cemented carbides, PKD has the capacity to machine workpieces at twice the speed and achieve a 10- to 20-fold increase in tool life in the process. The machining temperatures, however, must be kept under a certain limit value in order to prevent a sudden change from mechanical to thermal wear. It is recommendable to use compressed air or minimum quantity lubrication as a coolant/lubricant. PKD tools have a great potential for milling applications with GJV, with the cutting temperatures generally much lower than with continuous cutting. Since PKD cutting tool materials have a higher fracture resistance and strength than ceramics, it is recommendable to apply neutral to positive geometries to PKD tools. This can reduce the resultant force and contact zone temperature, which in turn allows for the application of higher cutting speeds [Pret06].

7.5.2.3 Cast Iron with Spheroidal Graphite

In cast iron with spheroidal graphite, also referred to as spheroidal cast iron, graphite is present in the form of spheroidal inclusions. In comparison to other cast irons with graphite inclusions, the base mass is less often interrupted by the graphite spheres and the internal notching is almost nonexistent. This increases the strength, with spheroidal cast iron reaching a high ductility [Herf07b].

The mechanical properties of this material, such as tensile strength and toughness, are determined by the ratio of ferrite to perlite in the matrix. Types with low strength and good toughness qualities (e.g. GJS 400-15) consist predominately of ferrite. High strength in combination with low toughness is typical for types with a predominately perlitic matrix (e.g. GJS 600-3). This ratio of ferrite to perlite is the result of the amount of bonded carbon in the matrix, which can be altered by means of a heat treatment. The machinability is also influenced by the perlite-ferrite ratio in the material matrix.

With an increasing perlite content, the strength of the cast iron rises along with the abrasive tool wear. The result of this is low tool life [Stau84]. Particularly suited to machining cast iron with spheroidal graphite are uncoated cemented carbides (HW), coated cemented carbides (HC) and oxide ceramics (CA) of the main application group K.

The resultant forces and the resultant effective mechanical stress of the tool cutting edge are relatively low when machining the GJS cast iron. Interrupted cuts can be used without difficulty because of the good damping property of the material. Increasing the cutting speed lightly reduces the values of the individual resultant force components.

Surface roughnesses of just under $R_a = 1\ \mu\text{m}$ are achieved in the finishing process. On a microscopic level, the surface is characterized by disembedded and in part, lubricated graphite inclusions. This effect only marginally degrades the measured roughness values.

Chips formed when machining cast iron with spheroidal graphite are, first and foremost, segmented chips. Only with very sharp cutting edges does a continuous chip develop which turns into a segmented chip given even a small cutting edge fillet. Helical chips may form which are slightly brittle because of the reduction in chip strength due to graphite inclusions. Because of the high temperatures in the contact zones between the workpiece and the tool, the material plasticises and is discharged between the workpiece cut surface and the flank face or between the lower side of the chip and the rake face. This phenomenon, referred to as the growth of built-up edge fragments, can occur when dry cutting at high cutting speeds.

In comparison to steels of the same hardness and similar strength, cast iron with spheroidal graphite must be machined at somewhat lower cutting speeds, especially when fine-machining with cemented carbides. Higher feeds, optimized tools and fewer steps as a result of a smaller material allowance can compensate for this disadvantage [Schm07].

A special heat treatment in the bainite stage can produce an austenitic/ferritic (also called ausferritic) matrix in cast iron with globular graphite. The term ADI (austempered ductile iron) is common for this. Figure 7.26 shows the structure of ADI. It consists of acicular ferrite and a high carbon-containing stabilized austenite. The material was erroneously referred to as bainitic cast iron in older publications due to the visual similarity of its structure with the ferritic carbide structure of steel. In addition to its increased strength and hardness, the matrix is characterized by increased toughness (e.g. GJS 900-7).

The special combination of abrasive and adhesive wear behaviour when machining ADI leads to extensive crater wear near the cutting edge, which leads in turn

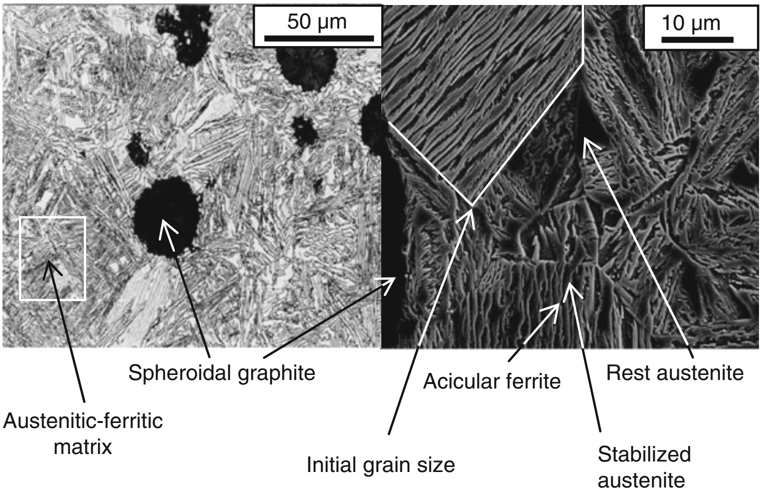


Fig. 7.26 Crystalline structure of austenitic-ferritic cast iron with speroidal graphite (ADI)

to a destabilization of the cutting edge. Therefore, the weak edge breaks before the critical flank wear will be reached. The use of a cutting fluid is recommendable for continuous cut operations. Mostly, coated cemented carbide tools of the main user group K, cBN and cutting ceramics are used as cutting tool materials [Kloc03].

The average forces arising when machining ADI are comparable to those of other cast irons with globular graphite. Due to its particular finely-striped grain structure, the average cutting forces are superimposed with a high dynamic. Surface quality is comparable to that of conventional cast iron with globular graphite. In the rim zone, the austenite converts into martensite because of mechanical stress, which is accompanied by an approximately 200–300 HB increase in hardness. Segmented chip formation is similar to that of conventional cast iron with globular graphite. ADI's high fracture strain has a negative effect on this [Klöp07].

ADI is basically not more difficult to cut than higher-strength GJS types. However, cutting strategies specially adjusted to the material are required. Lower cutting speeds due to high amounts of heat release can be compensated by increased feeds. Tool geometries specially tuned to ADI machining, such as drills with a radius or facette shape have a longer tool life. The use of cutting fluids is beneficial in continuous cut operations. For discontinuous processes such as milling, cutting fluids are not recommendable due to thermoshock stress on the tool [Schm07].

7.6 Machinability of Non-ferrous Metals

7.6.1 Aluminium Alloys

Mentionable areas of application of aluminium alloys as constructional materials include the following industries:

- automotive engineering
- aerospace engineering
- installation and apparatus engineering
- electrical engineering
- food technology
- chemical industry
- optical industry

From the realm of aeronautic engineering, the example of a modern commercial aircraft, which consists of up to 65–81 mass percent aluminium, makes it clear how many aluminium components are used [Star96, Töns01]. Aeronautic engineering traditionally uses a large amount of aluminium components, which are often characterized by a large percentage of machined semifinished products. In the chip-removing manufacture of structural components, this percentage can be in the order of 90–95% [Schu96, Berk01].

Aluminium alloys are distinguished by a low density of $\rho_{Al} = 2.7$ kg/dm^3 but also by low strength. In order to expand the use of aluminium alloys, obviously their strength must be increased. Alloy technology makes this increase possible with the main alloying elements manganese, magnesium, silicon, zinc and copper. A distinction is made between wrought alloys and casting alloys. Alloys of both groups can exist in both a self-hardening or hardened state. In the case of wrought alloys, the semifinished products are manufactured by forming. Here, the properties of the formed structure are decisive for machining. In the case of hardening (hot age hardening or room temperature precipitation hardening), strength is increased by depositing hard structural components, which generally are located on the grain boundaries.

The intermetallic phases should also be mentioned as another important group, the properties of which are very different from those of its components. Materials whose structure primarily consists of intermetallic phases are often abbreviated simply to "intermetallic phases" in common usage. NiAl, Ni_3Al and TiAl are key examples.

The machinability of aluminium alloys depends on their composition and structural state. Compared with cutting steels, boundary surface temperatures are much lower, around 350°C. Due to the low melting point of aluminium alloys, one must take care that the boundary surface temperature does not get too close to the melting point. The boundary surface temperature is very much contingent on the tribological conditions. Because of its low strength compared with steel, lower mechanical and thermal stresses on the cutting edge are to be expected for the same cutting parameters. The low boundary surface temperature makes it possible to use higher cutting parameters than in steel machining, most importantly higher cutting speeds, which are delimited from below by adhesive wear and from above by temperature resistance of the material.

The tool life parameters of machining tools used on aluminium alloys vary depending on the alloy and the structural state. Generally, more favourable tool life parameters can be expected than in steel machining. Adhesion is mostly dominant in the lower cutting speed range, although, as mentioned already in Chap. 3, abrasion is always active as well, increasing especially with a larger distribution density of hard particle inclusions. Such particles can be intermetallic compounds and non-metallic inclusions such as impurities of the molten bath. Hardened wrought alloys and casting alloys with a silicon content of up to 12% cause increased tool wear with increasing amounts. Due to the higher boundary surface temperatures, plasticized material can escape between the cut surface and the flank face and/or between the chip and the rake face. This phenomenon contributes to the deterioration of surface quality. The cutting speed must be reduced in these conditions. Cutting tools made of uncoated cemented carbide (HW) and diamond (DP) are the most commonly used cutting materials for machining aluminium alloys. Cemented carbides are used as cutting tool materials for machining wrought alloys and sub-eutectic casting alloys due to their wear resistance and hardness. In contrast to diamond, here the main stress is on toughness and the ability to manufacture complex tool geometries such as highly twisted end mill cutters with sharp edges. Diamonds, in polycrystalline form (DP) and in monocrystalline form (DM), are the first choice for machining strongly abrasive super-eutectic casting alloys. For all aluminium alloys, the use of cutting tool materials of the material group DP is not to be recommended for drilling into solid blocks due to compression processes in the area of the chisel edge. In the case of boring, especially aluminium alloys with high amounts of silicon, cutting tool materials of the material group DP are superior tot hose of the HW group with respect to tool life parameters and material removal rate. All cutting tool materials for main application group K can be used sensibly. Selection is based on general criteria such as cutting speed, cross-section of undeformed chip, and continuous or interrupted cut [Bech63, Opit64b, Zoll69, Bömc87]. Cutting trials have shown that CVD diamond coatings on cemented carbide substrates have the potential to combine the advantages of cemented carbides with those of polycrystalline diamond. When machining the wrought aluminium alloy AlCu4Mg1 (2024), the CVD diamond coating makes it possible, for example, to improve the tool life in comparison to uncoated tools. In interrupted cut with low cutting speeds and high material removal rates, high-speed steel (HSS) can be used advantageously to machine alloys with small amounts of silicon.

The specific resultant force of the types AlMg5 (5019), AC-AlSi6Cu4 (AC-42000) and AC-AlSi10Mg (AC-470000) is about 25% below those of the heat-treated steel C35.

Diamond cutting tool materials are often used to create highly reflective surface properties. Generally, these processes use high cutting speeds and small cross-sections of undeformed chip. The surface quality depends to a great extent on the wear mechanism of adhesion, which manifests itself in built-up edges and can be influenced by the process kinematics. Adhesion prevents the attainment of an optimal surface quality.

When there is a high material removal rate in an aluminium alloy machining operation, a large amount of chips has to be removed. For the sake of undisturbed manufacture, the chip form is a particularly important criterion for judging machinability.

The chip form can be affected by the alloy components, by heat treatment and by the process kinematics. When machining non-hardenable and hardenable alloys in a soft state without any percentage of silicon, long ribbon chips arise which make the machining process difficult. These alloys should be avoided as much as possible for components requiring machining [Bech62]. In hardened wrought alloys and casting alloys with a silicon content of up to 12%, increased silicon content leads to a more advantageous chip breakage. Hard and brittle inclusions such as Al_2O_3 and silicon also benefit chip breakage.

In summary, it can be said that cutting aluminium alloys forms the more favourable chips the harder they are. The most difficult to cut are especially non-hardenable aluminium alloys and hardenable aluminium alloys in a soft state. For this reason, it is recommendable if possible to cut these alloys in a condition of increased strength (cold-twisted/hardened) [John84].

7.6.2 Magnesium Alloys

Magnesium alloys have been used in Germany since the end of the 1930s especially in automotive engineering as light constructional materials in Volkswagen manufacture. Approximately 80% of the cast pieces produced in Germany currently consist of magnesium alloys [Wolf03]. Examples include transmission housings, steering parts, seat frames, inner linings, instrument supports and engine mounts in magnesium-aluminium composite structures [Dörn04]. Other products using magnesium include manual electronic devices, laptop housings, mobile phones and other devices in the area of consumer electronics as well as machine parts and sports articles. In comparison, sand casting is of only minor importance. The same is true of products made of magnesium wrought alloys. It has become clear that magnesium alloy machining is focused primarily on die-cast components.

The low density of $\rho_{Mg} = 1.74\,\mathrm{kg/dm^3}$ in comparison with $\rho_{Al} = 2.7\,\mathrm{kg/dm^3}$ and steel with $\rho_{steel} = 7.8\,\mathrm{kg/dm^3}$ has predestined magnesium alloys for applications in lightweight construction in particular. Alloy engineering aims to improve the strength properties of magnesium, especially for applications at temperatures above ca. 100°C. Another important goal of alloy technology is the improvement of the corrosion resistance of magnesium alloys, with respect to both inner corrosion and contact corrosion. The most important alloying elements include aluminium, zinc, manganese, silicon and rare earth elements such as cerium, lanthanum, neodymium, praseodymium and yttrium. Of the many types of magnesium alloys, the following will take a closer look at the example of one alloy group.

The alloy AZ91hp is the most frequently utilized magnesium alloy for die-cast parts. The notation “hp” (high purity) means that only trace elements of iron, copper and nickel are found. These elements cause inner corrosion in magnesium alloys.

AZ91hp is characterized by good mechanical properties and very good castability [Kamm00, DINEN1753]. Solidifying the alloy in a temperature range of 465–598°C favours inhomogeneous crystalline structures in the cast piece and the formation of porosities. This can impair the density of cast pieces, especially after machining.

Magnesium-aluminium-manganese alloys (AM-alloys) have very good strength properties together with higher fracture strain and notched-bar impact work. The corrosion resistance of alloys of this AM group is better than that of group AZ. The alloying element manganese bonds corrosive elements like iron, copper and nickel. Manganese and aluminium combine to form hard MnAl deposits, which accelerate tool wear. Because of their good castability, AM-alloys are used in automotive engineering, e.g. for structural parts, steering parts and rims.

AS-alloys (magnesium-aluminium-silicon) have similar mechanical parameters as AM-alloys. They are inferior however with respect to toughness. Their particular advantage is improved creep strength for temperatures up to about 150°C. Further alloy systems that aim to improve mechanical properties and creep strength include:

- Magnesium-zinc-RE-Zr (ZE-group)
- Magnesium-RE-Ag-Zr (QE-group)
- Magnesium-yttrium-RE-Zr (WE-group)
- Magnesium-aluminium-RE (AE-group)

The rare earth elements (RE) include cerium, lanthanum, neodymium, praseodymium and yttrium. The alloy AE42 contains RE as a misch metal with the composition 50% cerium, 25% lanthanum, 20% neodymium and 3% praseodymium. With the exception of higher fracture strain and higher creep strength, it has similar attributes as the alloy AS41. In order to classify material properties into a material spectrum, fracture strain at room temperature is a helpful mechanical parameter ($A = 7-14\%$ for magnesium alloys).

Beyond a certain temperature, about 225°C in the case of magnesium alloys, material behaviour shifts from brittleness to toughness. This phenomenon can be explained by the material structure. Magnesium has a hexagonal lattice structure at room temperature. Only the base planes act as slide planes at room temperature. Above $\vartheta = 225°C$, it changes into a cubic lattice structure. This increases deformability.

When cutting magnesium, lamellar chips are formed (Fig. 7.27). The distance between the lamellae or frequency of chip formation depends essentially on the tribology of the boundary surface, which can in turn be affected by feed and cutting speed. Characteristic of magnesium alloy cutting is the marked stick-slip friction, which is indeed contingent in a certain way on the material-cutting tool material combination, but can also be influenced by process kinematics. This means that a smart choice of process kinematics can adjust the amplitude and frequency of dynamic stress on the cutting edge to its loadability. Near the tool-chip boundary surface, friction causes temperatures to increase, plasticizing a thin material layer which holds the individual chip lamellae together.

Sharp tool cutting edges and smooth rake faces lead to good machining results. The tool geometry is based on those of tools for machining aluminium (Sect. 7.6.1). In drilling and milling with small tool diameters, the maximum spindle rotation speed puts a limit on the cutting speed. When rotating tools with large diameters are

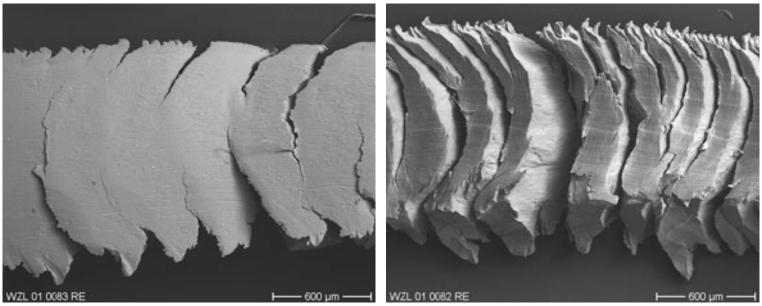


Fig. 7.27 Chip of the magnesium alloy AM60

used, the maximum allowable tool rotation speeds are the limit criteria. Workpiece stiffness can limit the chip cross-section and material removal rate, especially in finishing.

With respect to machinability, magnesium alloys are characterized by the fact that they contain few abrasive components. This is also true for the rim zones of the pieces to be machined, since these are predominately manufactured by die casting. When cutting magnesium alloys have only a slight tendency to adhesion, so no built-up edge formation is to be expected. The melting point of this alloy is in the range of 420–435°C. This makes it clear that the thermal load on the tool is relatively low in machining magnesium alloys.

Fitting cutting tool materials for machining include high speed steel (HSS), uncoated and coated ultrafine-grain cemented carbide (HF, HC), polycrystalline diamond (DP) and diamond-coated ultrafine-grain cemented carbides. In practice, ultrafine-grain cemented carbides of the application group N10/20 and polycrystalline diamond (DP) are the usual cutting tool materials. Tools made of these materials allow for high cutting speeds and feeds. They are characterized by especially high wear resistance and contribute to a great extent to process safety. The resultant force is relatively low in magnesium machining, below those of sub-eutectic aluminium alloy machining.

Figure 7.28 compares experimental results from dry face milling aluminium alloy AlSi9Cu3 and magnesium alloy AZ91hp [Kloc00]. Both alloys are used, for example, for transmission housings. When drilling both of these alloys with tools made of cemented carbide, the resultant force of the magnesium alloy is also below that of the aluminium alloy. Differences in cutting momentum are marginal [Wein00].

The possibility of machining magnesium alloys with high and extremely high cutting speeds is a good foundation for creating high surface qualities on machined components. Milling produces satisfactory surface qualities both in wet cutting when oil or emulsions are used and in dry cutting. In the case of the cutting

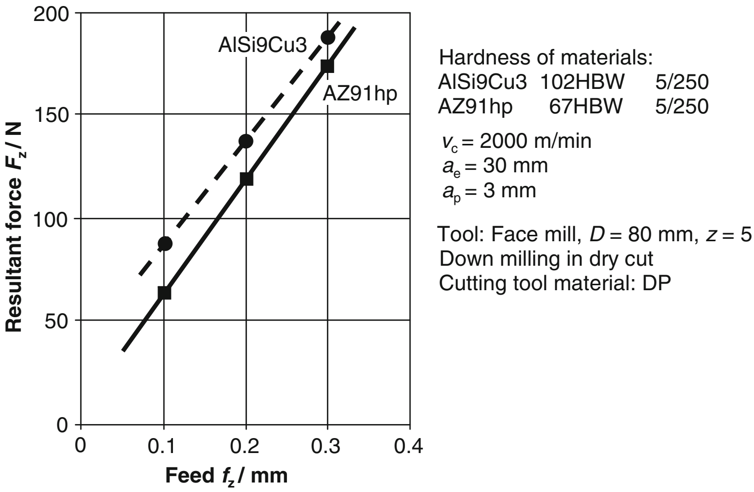


Fig. 7.28 Comparison of force components during face milling (dry cut)

conditions given in Fig. 7.28, an arithmetical average roughness of $R_a = 0.6\,\mu m$ was measured after wet face milling. For seal faces of motor vehicle components, R_a-values of about 1.5 μm are often required. These values can be safely adhered to with feeds of $f_z = 0.3$ mm [Kloc02]. In drill hole finishing, top surface qualities in the range of $R_a = 0.2 - 0.4\,\mu m$ can be safely obtained under production conditions, even when the highest cutting parameters are used [Wein00]. To achieve this, the process must be stable.

Due to the formation of lamellar chips, longer chips are almost completely absent, mostly spiral chip segments and discontinuous chips occur. Longer chip pieces break into smaller pieces when they make contact with the tool, the workpiece and the cutting medium because of their low stability. In general, the chip form poses no problem in magnesium machining.

7.6.3 Titanium Alloys

Titanium alloys are still a relatively recent group of construction materials. The first alloys were developed at the end of the 1940s in the USA, including the classic alloy TiAl6V4.

With a density of $\rho_{Ti} = 4.51$ kg/dm^3, titanium is the heaviest element in the light metal group. The great technical importance of titanium alloys is based above all on their high strength, but in particular on the ratio of yield strength to density, which is not even closely approached by any other metallic material (Fig. 7.29). Even high-strength steels with yield strengths of ca. 1000 Mpa are still little more than half that of a TiAl6V4 titanium alloy with respect to this ratio. Another important property of titanium alloys is their good corrosion resistance. They are resistant to temperatures of about 550°C. The maximum operating temperature is limited by the increase

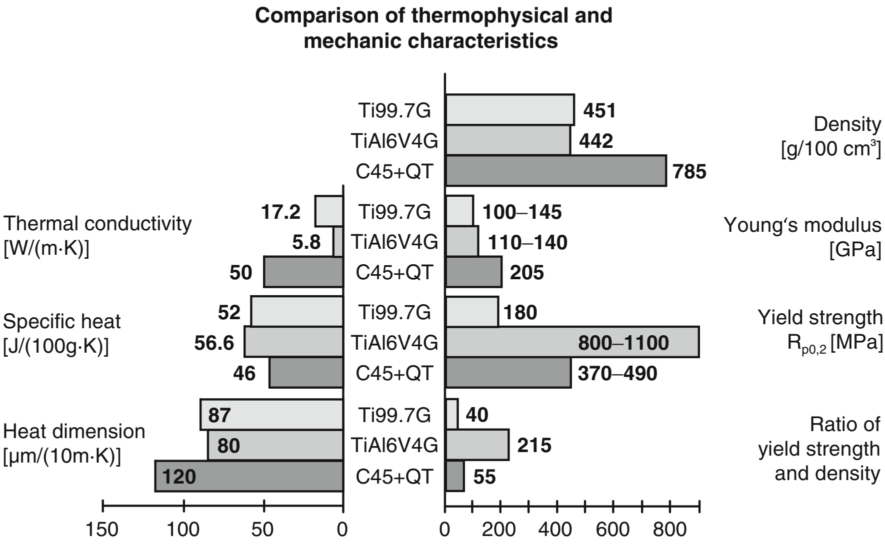


Fig. 7.29 Mechanical and physical properties of titanium materials in comparison to the heat treatable steel C45+QT (reference data)

in oxidation beyond 550°C [Essl91]. Also worthy of mention is the excellent biocompatibility of titanium alloys. Titanium alloys are used above all in these areas:

- aerospace industry
- chemical industry
- medical technology
- power engineering
- sports equipment technology

Titanium appears in various lattice modifications, of which each is only stable within a certain temperature range. At lower temperatures, pure titanium and most titanium alloys exist in the modification of a hexogonally close sphere packing, which is called α-titanium. The high-temperature phase on the other hand has cubic body-centred crystal lattice and is called β-titanium. The conversion of the α-phase into the β-phase takes place above 822°C, the "transus temperature". The existence of these two phases and the associated structural transformation is the foundation for the diverse properties of titanium alloys [Pete02].

In accordance to their influence on transus temperature, a distinction is drawn between neutral (Sn, Zr), α-stabilizing (AL, O, N, C) and β-stabilizing (Mo, V, Ta, Mn, Cr, Cu, Fe) alloying elements. The α-stabilizing elements – which include aluminium, the most important alloying element of titanium – extend the α-phase zone to higher temperatures and form a two-phase (α + β)-zone [Pete02].

The properties of titanium alloys are basically determined by their chemical composition and structure. The chemical composition primarily determines the parts by volume of the α und β phases. Structure is defined here in the case of almost exclusively two-phase titanium alloys primarily as the size and arrangement of both of these phases. The two extreme forms of phase arrangement are the lamellar structure, which arises by simple cooling from the ß-zone, and the globular structure, which results from the recrystallization process. Both structural types can exist in a fine or coarse form. Basically, the always varying structure is created by means of thermomechanical treatment. This involves a complex sequence of solution heat treatment, deformation, recrystallization, aging and stress-relief heat treatments [Pete02].

Titanium materials can be classified in the following groups:

- pure titanium
- α-alloys
- (α + β)-alloys
- near-α-alloys
- metastable β-alloys
- γ-titanium aluminide alloys

Pure titanium (e.g. ultrapure titanium: 99.98% Ti, pure Ti: 0.2Fe-0.18O or 0.5Fe-0.40O) and titanium alloys (e.g. Ti-5Al-2.5Sn, Ti-0.2Pd), which only contain α-stabilizing and/or neutral alloying elements, are designated as α-alloys. Because they are single-phase, α-alloys are relatively low in strength ($R_m = 280-740\,N/mm^2$). They are primarily used in the chemical industry and process engineering, since what is of chief interest here is good corrosion resistance and deformability. Pure titanium types contain up to 0.40% oxygen, which, as an interstitial alloying element, drastically increases the yield strength. Whereas oxygen is the only element intentionally added in order to meet strength requirements, there are other elements, such as iron or carbon, which represent impurities caused by the manufacturing process. The high level of corrosion resistance is achieved by alloying palladium (up to 0.2%). These titanium types are generally used in low temperatures. If higher strength is required, titanium materials alloyed with aluminium and tin (TiAl5Sn2,5) are also available [Pete02, Schu04].

The microstructure of (α + β)-alloys is characterized by a bimodal structure. It consists partially of globular (primary) α-phase in a matrix of lamellar α- and β-phase. This microstructure combines the good properties of lamellar structures, such as high resistance to creeping and fatigue crack growth, with those of globular structures, such as higher strength and fracture strain. The (α + β)-structure is fabricated by thermo-mechanical treatment, consisting of a combination of forming and heat treatment. By hardening, (α + β)-alloys reach very high strengths in the range of R_m = 900–1300 N/mm^2. Moreover, these alloys are characterized by high temperature resistance. Among the (α + β)-alloys is the by far most common titanium alloy TiAl6V4 (Ti-6Al-4V). More than half of all titanium materials are melted with this composition. It was already developed in the early 1950s in the USA and is both the most researched and tried titanium alloy. It is used above all in the aerospace industry. Titanium alloys are used to manufacture guide blades, impeller blades, plates, housings, pipes and distance rings between the rotor stages [Adam98, Pete02, Schu04].

Near-α-titanium alloys (e.g. Ti-6Al-2Sn-4Zr-2Mo-0.1Si) are the classic high temperature alloys, used in operating temperatures of 500–550°C. This alloy class combines the good creep properties of α-alloys with the high strength of (α + β)-alloys. Also, they exhibit a favourable combination of mechanical properties and oxidation resistance. These alloys have become very important in engine construction. They make it possible to increase operating temperatures up to 600°C. While the standard (α + β)-alloy TiAl6V4 can be utilized only in the relatively cold temperature range of the compressor, near-α-alloys can also be used in hotter areas, e.g. in high-pressure compressors. Two essential representatives of this alloy group are the titanium alloys Ti-6-2-4-2 (Ti-6Al-2Sn-4Zr-2Mo) and IMI834 (Ti-6Al-4Sn-3.5Zn). Both alloys have a bimodal structure (Fig. 7.30), however with a much higher percentage of α-phase compared with the standard alloy TiAl6V4.

Metastable β-alloys (e.g. Ti-4, 5Al-3V-2Mo-2Fe, Ti-15V-3Cr-3Al-3Sn) contain vanadium, molybdenum, manganese, chrome, copper and iron. Vanadium and molybdenum form a continuous series of mixed crystals with titanium, which remain stable at low temperatures. The mixed crystals decompose with the other alloying elements eutectoidically at low temperatures. Metastable β-alloys can be hardened to extremely high strengths of over 1400 MPa. Their complex microstructure allows for an optimization of the ratio of high strength to high fracture toughness. These alloys are used to produce foils, aerospace components and automobile components. Their use is limited however by their larger specific weight, moderate weldability, poor oxidation properties and the complexity of their microstructure [Pete02, Schu04].

Another material group based on titanium are the titanium alumnide alloys (α_2-Ti_3Al, γ-TiAl) first developed in the 1970s. Titanium alumnides are counted among the "intermetallic" phases. These are defined as compounds of metals with metals or non-metals, which have a different lattice structure than the initial component, from which their specific properties result. Intermetallic phases of the TiAl

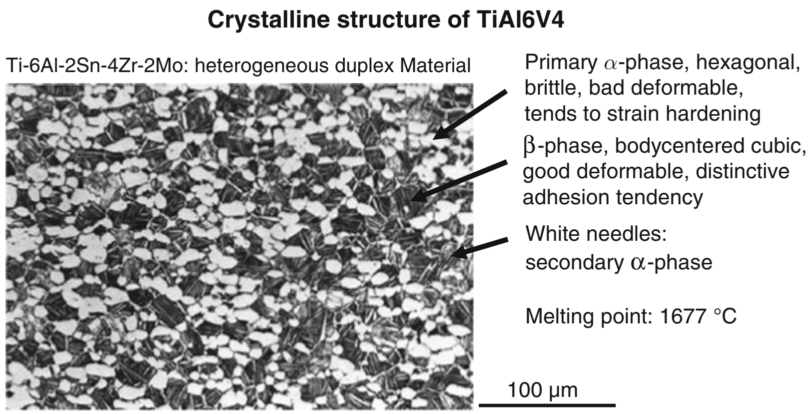


Fig. 7.30 Typical (α+β) mixed structure as in a compressor blade made of the near-α-alloy Ti-6-2-4-2, acc. to ADAM [Adam98]

type are characterized by low density, high high-temperature strength, high oxidation resistance and creep strength up to temperatures of 650°C. They thus are more temperature-resistant than the above-mentioned titanium alloys and are in direct competition with established nickel alloys with high density, which can be used in temperatures up to about 700°C [Pete02]. γ-titanium alumnide alloys are also characterized by their very low fracture strain at room temperature, in the area of $A = 0.5–1\%$.

Titanium alloys are difficult to machine because of their mechanical and physical properties. Their strength is high, and their fracture strain ($A_5 = 5–15\%$) is low. Their Young's modulus is almost half that of steel. The hexagonal α-phase is relatively hard, brittle, poorly deformable and has a high tendency towards strain-hardening. This phase affects the active tool cutting edge like the strongly wearing cementite lamellae in the perlite grains of carbon steels. The cubic body-centred ß-phase is very similar in its machinability to ferrite, which also crystallizes into the krz lattice type: it is easily deformable, relatively soft and ductile and has a high adhesive tendency (Fig. 7.30).

One important physical property for the machinability of titanium alloys is their low thermal conductivity, which is only about 10–20% that of steel (Fig. 7.29). As a result of this, only a small amount of the arising heat is removed with the chip. In comparison to machining the steel material C45E, about 20–30% more heat must be absorbed by the tool, depending on the thermal conductivity of the cutting tool material, when the titanium alloy TiAl6V4 is machined (Fig. 7.31). The result of this is that the cutting tool is subjected to high thermal stress, much more than is the

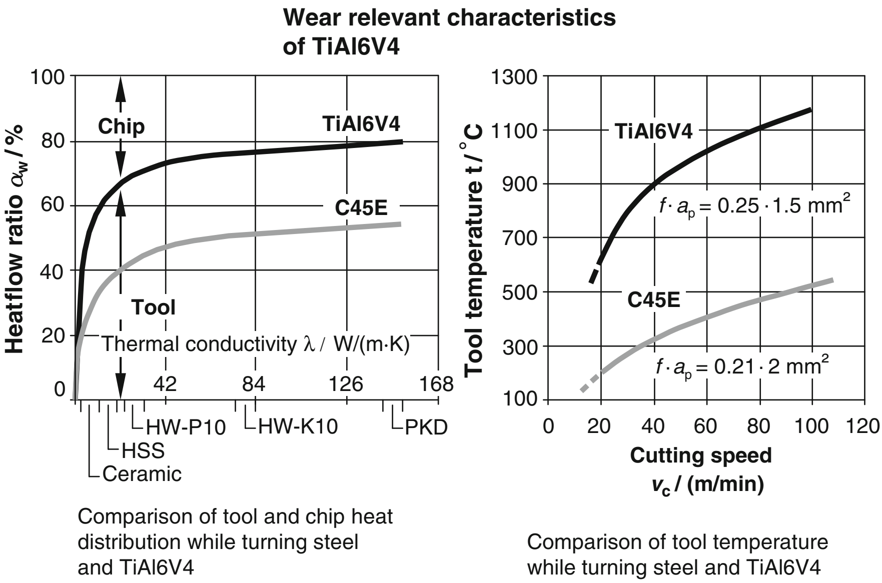


Fig. 7.31 Heat distribution coefficient of heat flow on the rake face and the thermal tool stress during turning of TiAl6V4 compared to C45E steel, acc. to KREIS [Krei73]

case when cutting steel (Fig. 7.31, right). This means that cutting titanium alloys not only exposes the cutting tools to considerable mechanical strain, but also to extremely high thermal stress [Krei73].

Another characteristic of titanium alloy machining under conventional cutting conditions is the formation of lamellar chips. The cause of this is a constant shift between compression and sliding phenomena in the shear zone (Fig. 7.32). In position I, the shear zone is already fully formed. The lamella slides over the rake face while a new lamella is compressed. Since the deformation resistance of the titanium material is rapidly decreased immediately after the shear zone is formed as a result of the high shear/deformation speed, the cutting force is steadily reduced. The simultaneously introduced compression of the newly forming lamella causes the cutting force to go up again (position II), until the cutting force or shear force is so high in position II that the shear strength of the material is exceeded and a new lamella is formed [Krei73]. Such discontinuous chip formation subjects the tools to a mechanical and thermal alternate load, the frequencies and amplitudes of which depend directly on the cutting conditions. The dynamic cutting force can amount to about 20–35% of the static cutting force. Mechanical and thermal alternating stress can lead to tool fatigue and encourage tool failure due to the formation of cracks,

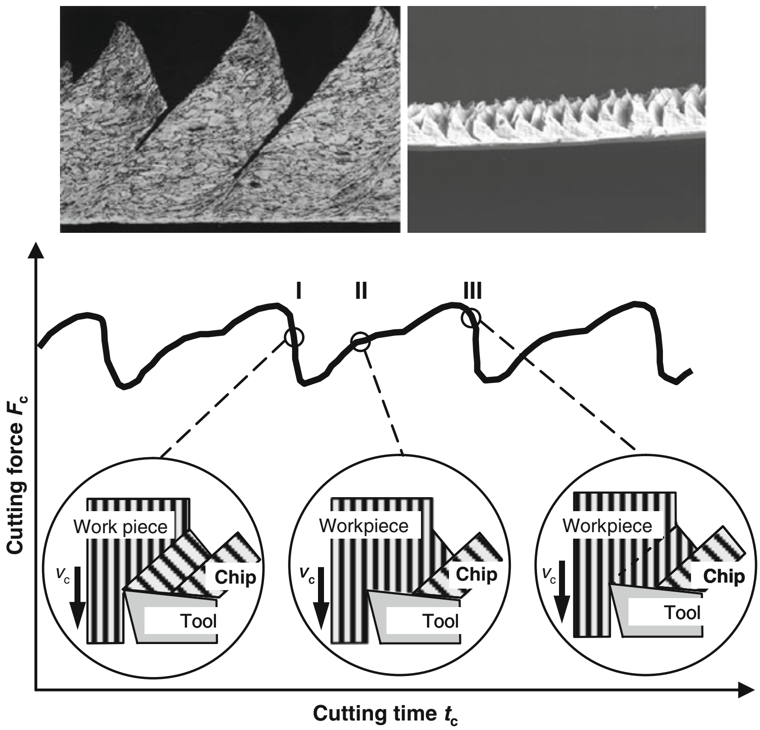


Fig. 7.32 Chip formation during turning of titanium, acc. to KREIS [Krei73]

shell-shaped spalling, the fracture of cutting tool material particles or cutting edge fracture [Krei73, Yang99, Kita97].

Titanium materials are generally turned with uncoated cemented carbides of the main application group S. Due to the high thermal and mechanical load, the tools can however only be used at relatively low cutting speeds. The range of usual cutting speeds is about 50–70 m/min for rough turning and 60–90 m/min for finish turning.

Cemented carbides containing titanium (P-types), cermets or coated cemented carbides are generally not suitable for turning titanium alloys. Elements of the substrate (Ti) or the finishing material (Ti, O, N) react with the workpiece material, which greatly reduces the wear resistance of the tools. In isolated cases, good results have been reported in the case of machining with TiB_2-coated tools.

Cutting tool materials based on Al_2O_3 and Si_3N_4 do not come into considerations for machining titanium materials due to their low thermal conductivity and the great affinity of aluminium, silicon, oxygen and nitrogen to titanium.

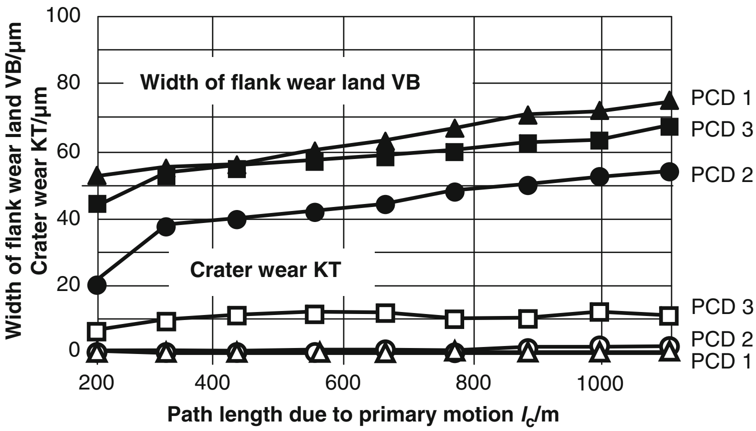
One alternative to uncoated cemented carbide tools in the case of finish turning titanium alloys are tools made of monocrystalline diamond, polycrystalline cubic diamond (PCD), CVD-diamond thick films and polycrystalline cubic boron nitride (PCBN). These cutting tool materials are characterized by high hardness and wear resistance, excellent thermal conductivity compared to other cutting tool materials (Fig. 7.31), low thermal expansion and low friction between the rake face and the chip (or between the flank face and the workpiece). Compared with cemented carbides, they make it possible to use higher cutting speeds, thus clearly reducing production time while maintaining the same or even improving cutting quality. The range of speeds applicable for tools made of these cutting tool materials is $v_c = 100-200\,\mathrm{m/min}$. Of the PCBN cutting tool materials, above all type containing high level of cBN are good for finish turning several types of titanium alloys.

When turning titanium alloys with PCD tools, the interactions that effect wear taking place between the workpiece material and the cutting tool material are extraordinarily complex. They are characterized by diffusion and graphitization, thermally caused cracking, surface damage as a result of lamellar chip formation and the potential formation of wear-reducing reaction films on the diamond grains [Bömc89, Neis94]. Due to these diverse interactions between the workpiece and cutting tool materials, the performance capacity of PCD cutting tool materials in the case of titanium machining is highly dependent on the composition of the cutting tool material. Especially mentionable in this context are above all the composition of the binder phase, its quantitative amount as well as the size of the diamond grains [Neis94].

The most dominant form of wear when turning titanium alloys with PCD tools is the formation of craters on the rake face. Flank face wear is of secondary importance, especially at high cutting speeds. In experiments where the titanium alloy TiAl6V4 was cut using external cylindrical turning, the lowest amount of crater wear was measured in the case of a PCD variety with SiC as binder. In the case of PCD types with cobalt-containing binders, it was seen that crater wear was greatly effected by the binder content and the grain size. The largest amount of crater wear was observed when using the PCD type with the largest quantity of binder and the smallest grain size [Köni93c, Neis94, Kloc07].

Due to its catalytic effect, cobalt promotes the graphitization of diamond. The result is that the cutting tool material becomes less resistant to abrasive wear. Furthermore, cobalt and diamond have different thermal expansion coefficients, which favours the formation of microcracks. This process can be observed above all in the case of fine-grain types. In conjunction with dynamic stress on the cutting tool material by lamellar chip formation, microcracks further the separation of individual diamond grains or entire grain bonds from the cutting tool material compound [Bömc89, Köni93c, Neis94] (Fig. 7.33).

The lower amount of crater wear when turning with SiC-containing or coarse-grain cobalt-containing PCD is attributed in the literature to the formation of a wear-inhibiting film made of titanium carbide on the diamond grains. It is believed that there is a reaction caused by diffusion between the titanium from the material and the carbon from the cutting tool material in the area of the crater. The TiC film that is formed in the area of the contact zone on the rake face adheres tightly to the diamond grains that make up the surface during the machining process. Since the speed of diffusion of carbon to titanium carbide is more than 10 times slower than that of carbon to titanium, the wear process is significantly reduced [Hart82]. Thus, to reduce crater formation when turning titanium alloys with PCD as much as possible, PCD cutting tool materials with large diamond grains, low amounts of cobalt or with a binder phase made of β-SiC should be used.



Composition	PCD 1 △▲	PCD 2 ○●	PCD 3 □■
Diamond *	92 %	92 %	80 %
Binder *	8 %	8 %	20 %
Grain size	6–10 µm	2–6µm	0.5–1 µm

Binder phase: WC-Co
* Volume percentage

Face turning, TiAl6V4
$v_c = 110$ m/min
$a_p = 2.0$ mm
$f = 0.1$ mm

Fig. 7.33 Diamond as cutting tool material for turning of titanium alloys

Weight reduction is an essential criterion for modern airplane engines. This can be realized by using materials of lower density, but also with the help of an innovative component design. One example of a modern component concept is the blade integrate disk or “blisk”. To manufacture a blisk, the blades must be carved from the solid material. This is an extremely demanding machining task, especially considering that the blades must meet the highest demands regarding surface quality, rim zone formation and formal/dimensional accuracy.

For blade lengths up to 80 mm, blisks are produced using end milling or form cutting. For blisks with a diameter of 600 mm and about 70 blades, up to 55 h of pure milling time is required. Cutting becomes progressively more difficult with increasing blade lengths and thus with increasing milling cutter protruding lengths. In such cases, ECM machining is preferred, though all the blades must first be pre-milled to an overmeasure of 3 mm. Here too, machining also requires up to 45 h depending on the size of the blisk, as well as 7 or 8 more hours should an ECM operation be included. Blisk production has thus proven to be very time-consuming and cost-intensive.

Blisk pre-milling is done with end milling cutters made of high performance high speed steel or conventional cemented carbides. Due to the large volume of material to be removed, tools with relatively large axial (D) and radial ($D/2$) depths of cut are used to realize large material removal rates. On the other hand, in order to finish the blades only very small overmeasures need to be removed. The main concern in this case is above all high process safety, surface quality as well as form and dimensional accuracy. For this type of machining, tools made of ultrafine-grain cemented carbide is recommendable due to its excellent wear and toughness properties.

Because of their high level of wear resistance and bending strength, higher speeds and larger material removal rates can be realized with end milling cutters made of ultrafine-grain cemented carbide than with ones made of conventional cemented carbide. In the example provided (Fig. 7.34), it was possible to reduce the total machining time required for blade finishing by about 50% by using ultrafine-grain cemented carbide milling cutters in conjunction with an adapted machining strategy. Cutting blades with these tools at high speeds lead moreover to still further advantages, the most important of which include improved surface quality of the milled blades and lower resultant forces, which make it possible to realize higher levels of component precision.

Trochoidal milling is one way to reduce milling time when pre-machining blisk blades (Fig. 7.47). Using this method, even curved blade surfaces can be manufactured in one cycle to nearly the final contour with end milling cutters on 5-axis machine tools by superimposing a wobbling motion.

Analogously to the remarks given above on milling nickel-based alloys (Sect. 7.6.5), the formation of the cutting edge has a key role with respect to tool wear when milling titanium alloys as well. Here too, results from research and industrial praxis have shown that stabilizing the cutting edge with a defined rounding in the order of 5–20 μm leads to a significant increase in tool life not only for long-protruding tools when end milling titanium alloys.

When milling materials that are difficult to machine, of which titanium alloys are an example, down milling should generally be given preference. In up milling, the

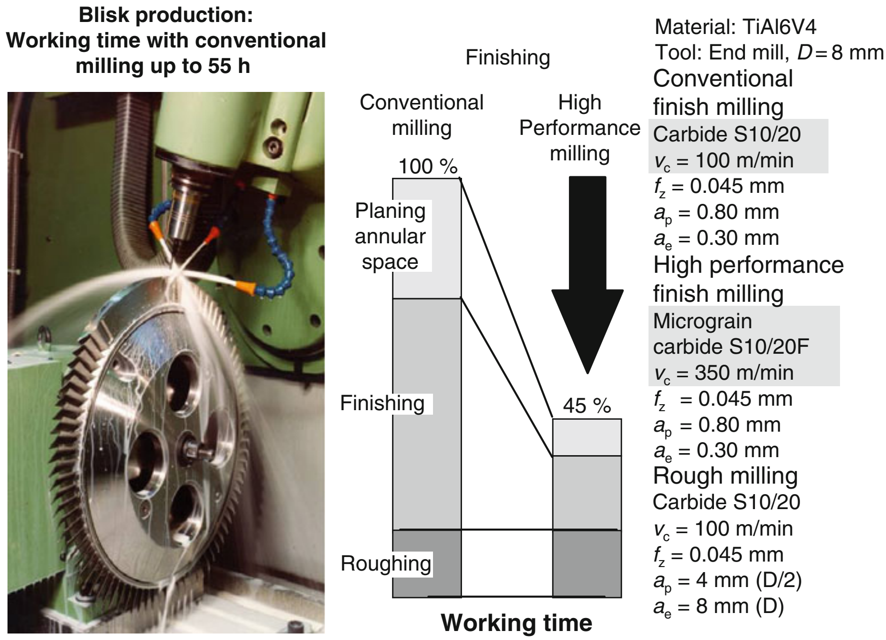


Fig. 7.34 Reduction of working time by applying high performance milling during finishing a blisk made of the titanium alloy TiAl6V4 (Source: MTU)

tool enters with a chip thickness of $h = 0$ mm. Due to the high elastic deformability of titanium alloys, chip formation is preceded by a long friction phase between the tool and the workpiece, which greatly promotes wear on the flank face. The tool exits with $h > 0$ mm except for the fluting from the solid. Chip root [Peke78] and tensile residual stresses can develop in the cutting edge as a result of this. Tensile residual stresses can lead to crack formation and then to local cutting tool material fracture and tool failure. Chips adhering to the cutting edge are partially compressed upon re-entry of the cutting edge into the material on the up milling flank, or they wind up between the cutting edge and the material. Faulty component surfaces and/or cutting tool material fractures on the cutting edge are the result.

In down milling on the other hand, the cutting edge enters with $h > 0$ mm and exits with $h = 0$ mm. Chip root formation is not possible when the tool exits the material. Potentially adhering chips are only connected with the cutting edge by a thin strip of material and are usually wiped from the tool upon its re-entry. The resultant surface quality of the down milling flank is much better than that of the up milling flank.

The special characteristic of down milling is that the cutting edge exits with a chip thickness of $h = 0$ mm. This is of great importance when milling grooves, also during the phase in which the tool enters the workpiece. If the tool enters in a straight line, the cutting edge always exits with $h > 0$ mm until the groove is fully cut, such as in up milling. If the tool enters on the other hand with an “arc lead”, the

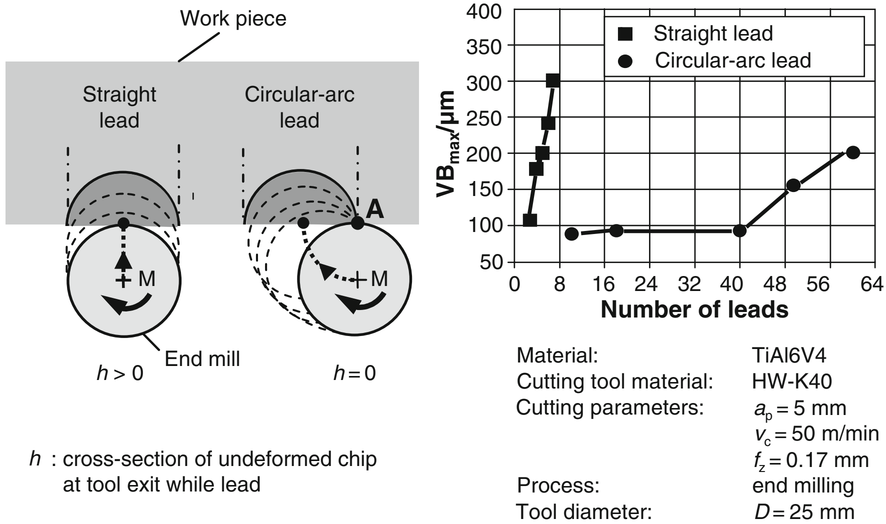


Fig. 7.35 Influence of the entering path on tool wear during milling the titanium alloy TiAl6V4

cutting edge will exit the material with $h \approx 0\,\text{mm}$, analogously to down milling. Both first cut strategies have a significant effect on the wear and performance of end milling cutters. Compared to the conventional lead, the arc lead results in much lower tool wear (Fig. 7.35).

In contrast to turning, coated tools can also be employed for many machining tasks when milling titanium alloys. Causes of the good wear and performance properties of coated tools in milling could include the lower tool temperature caused by the interrupted cut and the compressive residual stresses characteristic of PVD coatings.

Because of the high thermal stress on the tool, titanium alloys are usually machined using wet cutting. Due to the long engagement times, an intensive cooling is necessary, especially when turning. One extremely effective method in this context is supplying a cutting fluid under high pressure ($p > 80$ bar). In experiments in turning the titanium alloy TiAl6V4 with uncoated cemented carbides, the cutting fluid was supplied both conventionally with a pump pressure of 6 bar and under high pressure with 140 bar (Fig. 7.36). Under the conditions of high pressure lubricoolant supply the resultant tool life was higher by a factor of 2.3 (VB = 0.3 mm) compared with conventional cutting fluid supply. As this example shows, a cutting fluid jet supplied with high pressure can be used to improve not the tool life, but above all the cutting speed and thus the process's productivity [Gold07]. The cutting fluid jet supplied to the gap between the chip bottom side and the rake face not only cools the tool intensively, it also improves chip fracture. Instead of long ribbon and snarled chips, short-breaking chips are formed under the selected conditions, the removal of which from the chip formation location and the machine tool no longer presents any problems. Especially in grooving operations, high pressure lubricoolant supply

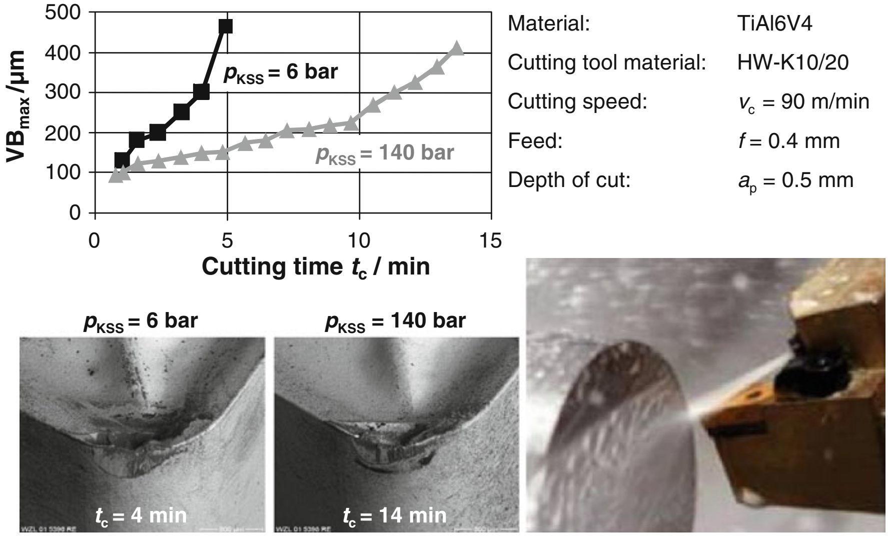


Fig. 7.36 Increase of tool life due to high pressure lubricoolant supply

can lead to enormous improvements in chip formation and tool performance. One method variant is supplying the cutting fluid through the insert. The cutting fluid exits near the cutting edge in the area of the flank or rake face.

The use of the high pressure lubricoolant supply is not limited to turning. It can also be used for drilling and milling operations. In the case of drilling, the cutting fluid is supplied via the spindle and the cooling ducts in the tool. The use of the high pressure lubricoolant supply when drilling into titanium-based or nickel-based alloys increases performance significantly. Compared with conventional supply, the resulting tool life is many times higher. While in the case of internal supply via the machine spindle, the pressure applicable through the pivoting feedthrough is currently limited to about 140 bar, in the case of external supply in turning or milling, the cutting fluid can be supplied with much higher pressure. At present, high pressure units with pressures of up to 1000 bar are available [Fili02]. Whether it is technologically necessary and economically sensible to work with such high cutting fluid pressures is currently being researched. As the first results have shown, cutting fluid pressures in the range of 100–200 bar should be sufficient for most machining tasks. In high pressure lubricoolant supply not only the pressure is important, but also the volume flow. The latter must also be optimally adjusted to the respective machining task.

Analogously to turning, cutting fluids are regularly used when milling titanium alloys as well. As in the case of milling steel materials, the basic problem is that the highly heated cutting edge emerging from the material is abruptly cooled down by the cutting fluid. The thermoshock caused by this promotes the formation of comb cracks and hence tool wear. In unfavourable cases, comb cracks, in conjunction with mechanical shock load of the cutting edge when entering the material, can lead to

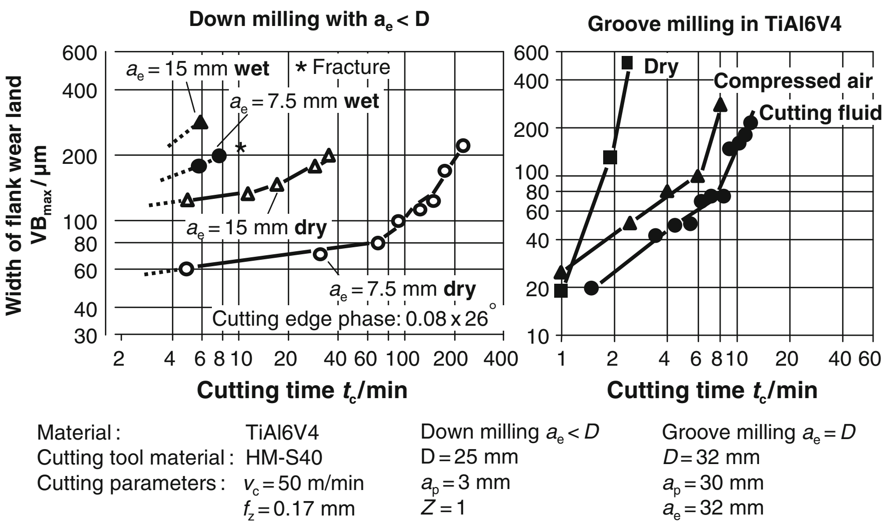


Fig. 7.37 Influence of wet or dry cutting conditions on the tool wear development during milling TiAl6V4

tool failure due to fracture. Numerous investigations into dry milling steel materials have proven that dispensing with the cutting fluids leads to a significant increase in tool life. In order to diminish thermal alternate stress and the resulting formation of comb cracks, the question is whether it makes technological sense to dispense with cutting fluids when milling titanium alloys too.

Comparative studies [Köll86] in down milling TiAl6V4 with $a_e < D$ have confirmed the general results gathered when dry milling steel materials (Fig. 7.37). In comparison to wet machining, dry cutting also leads here to a significantly lower amount of wear on the end milling cutters used. In the case of down milling with $a_e/D = 7.5/25$, the tool's life had ended already after 8 min with wet cutting, while in dry cutting its life was extended to 200 min.

This result changes fundamentally when milling in full groove cutting, i.e. as soon as the cutting edge penetrates the material with a chip thickness of $h = 0$ mm. In this case, the lowest amount of wear was recorded when milling with a cutting fluid. The reason why wear is increased when dry milling is that chips adhering to the cutting edge or found in the groove end up in the material between the cutting edge and the groove flank, are compressed on the surface of the up milling flank and cause cutting tool material fracture on the cutting edge. When milling in full groove cutting, the cutting fluid has the essential task of separating the chips from the cutting edge and transporting them out of the slot [Köll86].

In principle, it is also possible under certain conditions to dry machine titanium alloys. In most application cases in praxis however, cutting fluids are used, not the least because of the safe chip removal and therefore for reasons of higher process safety. In these cases, one should take care that cutting fluid supply to the tool is carefully adjusted in order to reduce thermoshock and that the tools are intensively cooled.

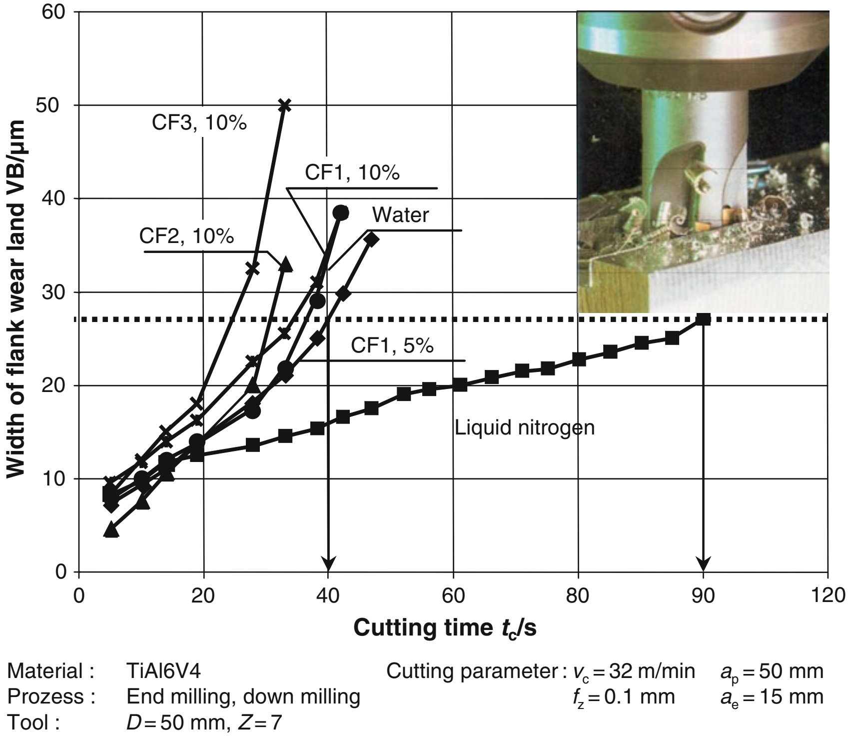


Fig. 7.38 Influence of cutting fluid on the tool wear development during milling TiAl6V4

One alternative to conventional flood cooling when milling is, in analogy to turning (Fig. 7.36), supplying the cutting fluid under high pressure. The use of cold or liquid gases is an extreme form of tool cooling. Investigations with various cutting fluid media, concentrations, water and liquid nitrogen had the following results (Fig. 7.38). There are at points significant differences in performance between the cutting fluid media. When milling titanium alloys, unsatisfactory results can also stem from the use of an unsuitable cutting fluid. Since tool cooling is especially important in titanium machining, the amount of oil in the emulsion should not be too high. In this investigation, a 5% emulsion lead to less tool wear than a 10% emulsion. As the more pronounced wear formation when milling with pure water containing only a rust inhibitor proves, a certain amount of oil must however be present in the cutting fluid in order to reduce friction.

In this comparison, cooling with liquid nitrogen delivered the best results (Fig. 7.38). The boiling point of liquid nitrogen is –195.8°C. The use of this cooling medium led to an extreme cooling of the tool and the chips. Similarly positive results can be obtained by the use of cold gases. One example of this is the use of CO_2-snow (dry ice) when turning a duplex steel. By means of the intensive cooling, tool wear and burr formation could be reduced in addition to chip formation and surface quality being improved [Wein07]. As these examples show, the use of extremely cold media can contribute to the solution of difficult machining tasks. Whether the cost associated with their use is economically worthwhile must be decided on a case-by-case basis.

Titanium alumnides may have the potential to raise the operating temperatures of titanium alloys to 800°C. They are for this reason a possible alternative to nickel-based alloys at about 50% the weight, not only for high pressure compressors but also for the low pressure turbine. Their low elongation at break (< 1% at ca. 700°C) and low thermal expansion ($\lambda = 10\,\text{W}/(\text{mK})$) make these alloys problematic for machining. The problems include high tool wear, extremely low applicable cutting speeds and insufficient surface quality. Components made of titanium alumnide have predominantly been machined with uncoated cemented carbides (HW-K10/20) [Ecks96, Aust99]. When turning however, tools made of PCD or with CVD thick diamond films can also be successfully employed.

The main problem in machining these materials under conventional process conditions is the formation of defects in the form of microcracks and micro-fractures on the surface of the machined workpieces (Fig. 7.39). The cause is the high level of brittleness of the material and its incapacity to deform plastically. This becomes very clear in consideration of the discontinuous and segmented chips that are formed. Chip formation such as it is familiar when machining steels or titanium alloys, is not found in conventional machining. Instead, angular, needle-shaped chip lamellae

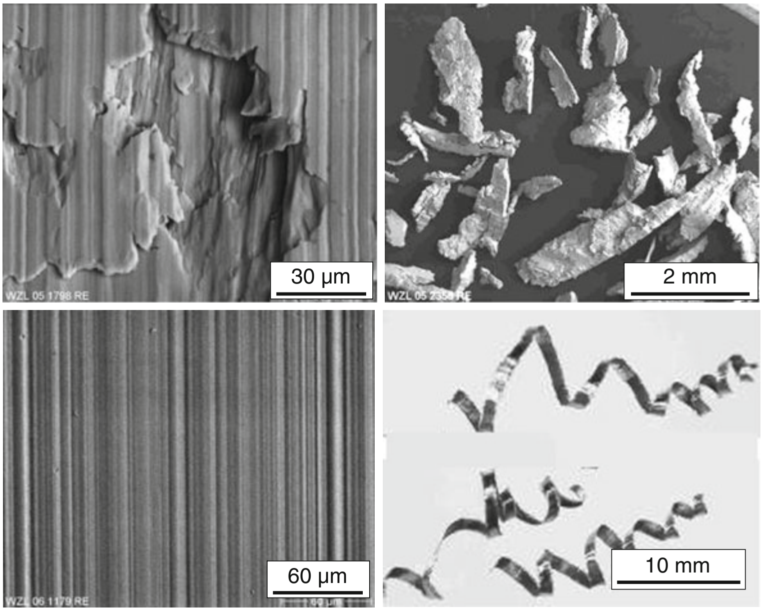


Fig. 7.39 Surface formation and chip forms during turning of γ-titanium aluminide with conventional and adapted cutting conditions

are formed whose separating surfaces are very rough, have obvious fracture structures and extensive cracks. Plastic deformation of the material during chip formation has not been observed. The material to be removed is, so to speak, pried out of the workpiece material. Such surface structures are unusable above all for components requiring high levels of safety such as are needed in engine construction.

As the basic research has shown, the machining result can be significantly improved by an adjustment of the cutting edge geometry and process parameters. With sharp-edged tools and the choice of a small ratio of depth of cut a_p to the corner radius r_ε, it is possible to produce smooth damage-free surfaces by turning workpieces made of Ti-45Al-8Nb-0.2C. Under these conditions, coiled chips are formed such as one finds in steel machining. When cemented carbides of the S10/20 group are used, cutting speeds of up to 100 m/min can be used.

7.6.4 Copper Alloys

Copper alloys are privileged because of their excellent thermal conductivity and corrosion resistance in the following areas:

- air-conditioning
- hydraulic engineering
- recuperator technology
- food technology
- chemical equipment and apparatus technology
- auxiliary equipment

Copper alloys are defined as alloys in which there is at least 50% copper. They can range from the superhard two-material system copper/aluminium to types of pure copper with low strength and high fracture stress. In comparison to other metallic construction materials, most copper alloys, as will be explained below, are considered easy to machine.

A distinction is drawn between wrought and cast materials. Within this classification, it is generally sensible to classify according to alloy groups. Although the properties of copper alloys are essentially determined by their chemical composition, this organizing principle is not suitable for classifying the machinability of copper alloys because of highly varying machinability of copper alloys of the same type [DKI83]. With respect to machinability, the following categorization is suggested:

- Pure copper and copper alloys with zinc, tin, nickel and aluminium, the additive elements of which are adjusted to each other such that they only form a homogeneous mixed crystal. Such alloys are easy to deform when cold and have a high level of deformability. This group is considered to be moderately to poorly machinable. Alloys of this group are especially familiar in the form of the two-material system copper-zinc, known as brass.

- Alloys with the elements zinc, tin, nickel, aluminium and silicon, however without chip-breaking additives, which form a second mixed crystal. Such heterogeneous alloys are harder than the previously mentioned group. They have less deformability than the previous group, but have higher machinability. This group is especially characterized by the three-material system copper/tin/zinc and copper/nickel/zinc, also known as nickel silver.
- Alloys of both above groups, to which are added lead, sulphur, selenium and tellurium as insoluble components in order to improve chip breakage. Strength is hardly affected, notched impact strength and deformability are reduced [Seid65]. This group is the most machinable because of its improved chip fracture. This group is made up of automatic alloys, to which the elements lead, tellurium or selenium are added so that an unproblematic chip breakage can reduce disturbances in the manufacturing process.

Special alloys can contain nickel, cobalt, tin or vanadium. Lead, bismuth, antimony and cadmium are contained in free cutting alloys as chip-breaking additives [John84]. Beryllium, boron and natrium are supplemented as trace additives in order to affect crystallization.

The emphasis of the main criteria of machinability changes depending on the machining task at hand, especially when machining copper alloys. For this reason, it is difficult to make a general classification of these alloys with respect to their machinability. However, there are a few essential points which influence the machinability of copper alloys that should be considered:

- the manufacturing process for producing semifinished products, e.g. primary forming or shaping
- heat treatment, e.g. hardening
- the chemical composition of the alloy

7.6.4.1 Influence of Semifinished Product Manufacture

When machining cast copper alloys, one must consider the structure of the rim zone, which is different from that of the core structure and is called casting skin in practice. This rim zone is characterized by a higher level of hardness and strength compared to the core structure, resulting in an acceleration of tool wear. The core structure of cast alloys is generally more machinable than that of wrought alloys [DKI83].

In the case of cold forming copper alloys, their hardness and strength are increased while their deformability is reduced. This leads to a positive influence on machinability due to improved chip fracture compared with materials that have not been shaped [DKI83].

7.6.4.2 Influence of Heat Treatment

Machining hardenable copper alloys is preferably executed with a cold-worked material prior to heat treatment, since machining after hardening would cause accelerated tool wear. On the other hand, the harder material state is preferable for subsequent grinding and polishing work.

7.6.4.3 Influence of Chemical Composition

Brass and pure copper (Group 1) are difficult to machine because of their high levels of toughness and deformability. They are characterized by high chip compression, which has a large impact on the tribology in the chip/rake face boundary surface and causes high mechanical stress on the cutting edge. In practice, these phenomena are known as “snagging”. The effect of the alloying elements lead, sulphur, selenium and tellurium on the form of insoluble components in copper alloys is comparable to the effects of these elements in automatic steel [Isle73, Lore74].

7.6.4.4 Wear

When machining tough alloys in the case of continuous chip formation and low temperature at the boundary surface between the chip and the tool, built-up edges can be observed that lead to accelerated wear of the cutting edge [Klei66, Mess69]. Due to hardness and deformability, the tool life parameters are less favourable when machining nickel silver than when machining brass [Vict72]. Built-up edges are formed in brass machining as a result of the wear mechanism of adhesion. In the case of high speed milling, uncoated cemented carbide of the application group K10/20 is recommended in order to guarantee good cutting edge durability. In the case of materials that are difficult to machine and tend to adhesion, such as pure copper or brass containing large amounts of copper, polycrystalline diamond (DP) has proven superior as a cutting tool material not only in terms of better wear resistance but also by its favourable tribological conditions, leading to improved surface quality and lower resultant forces. Ceramic cutting tool materials are not suited to cutting copper alloys because of their adhesive tendency [grei91].

7.6.4.5 Resultant Force

As the cutting speed rises the specific cutting force k_c falls in the case of copper alloy machining as well. When the cutting speed is increased from 5 to 160 m/min when performing an external cylindrical turning operation on copper, the specific cutting force is reduced, for example, by about a third [DKI83]. Further increase of the cutting speed causes the specific cutting force to converge asymptotically towards a constant value. It has become clear that, in the case of the cutting speeds commonly used today with cemented carbides as cutting tool materials (> 160 m/min), the influence of cutting speed can be neglected. Since the specific resultant force in the case of copper alloys is generally considerably lower than that in cutting of steel, difficulties due to insufficient driving power of the machine tool can hardly be expected in practice. For cast alloys, the casting method used to produce the semifinished products has a considerable effect on the amount of specific resultant force. Machining experiments have determined that with constant cutting values parameters the amount of specific resultant force is lower for centrifugal casting than for sand casting, although the centrifugally-cast workpieces had higher tensile strength, Brinell hardness and strain. Due to the finer crystalline structure of centrifugally-cast parts, chip formation also proved more favourable.

7.6.4.6 Surface Quality

Built-up edge formation and flank face wear of the minor flank face lead to poor surface quality [Klei66, Mess69]. In the case of thin-walled workpieces, deformations of the workpiece due to the cutting force can appear as a result of the low elastic modulus of copper alloys (e.g. CuZn30: 115.000 N/mm^2 with RT). These deformation not only endanger dimensional accuracy, but also induce undesired residual stresses in the rim zone. Lowering the cutting force can lead to improved quality. As a rule, the use of a cutting fluid improves the surface quality as well [Grei91].

7.6.4.7 Chip Form

The machinability of nickel silver (Group 2) can vary to a great extent depending on the respective amounts of the alloying elements zinc, tin, nickel, aluminium and silicon. Usually, acceptable chip forms are obtained however. The chip formation of pure and homogeneous copper is relatively unfavourable. Large cross-sections of undeformed chip and an unhindered chip flow promote long ribbon chips. With additional alloying elements (Pb, Te, S, Se), chip fracture can be significantly improved, and thus the chip form as well. For example, an alloy made of tellurium-containing copper CuTeP can even be machined on automatic machines because of the short-breaking chips. In contrast to unalloyed copper, the alloy CuTeP has only slightly less thermal conductivity [DKI83].

7.6.5 Nickel Alloys

Nickel alloys are materials with nickel as their main component, which are alloyed with at least one other element. They are characterized by good corrosion resistance and/or excellent high temperature strength.

Nickel has a cubic body-centred crystal lattice and, at low temperatures, also has excellent ductility and cold-formability. It is one of the ferromagnetic materials.

Some nickel alloys have special physical properties. One example is the soft-magnetic nickel-iron alloy with about 15% Fe, 5% Cu and 4% Mo, also familiar under the name Mumetall, which is used for shielding magnetic fields because of its high magnetic permeability [Schu04].

Nickel alloys are often applied as construction materials due to their superior properties. Examples include:

- chemical industry (cauldrons, heat exchangers, valves, pumps),
- environmental protection and waste management (flue gas desulphurisation plants),
- power production (power plant generators),
- aerospace (engines).

The specific properties, adjusted to the respective application area, are essentially dependent on the chemical composition, possible cold-shaping and the type of heat treatment. In accordance with their most important alloying elements, nickel alloys can be classified in the following main groups (Fig. 7.40) [DIN17742, DIN17743, DIN17744, DIN17745, Ever71]:

- I nickel-copper alloys
- II nickel-molybdenum alloys and nickel-chrome-molybdenum alloys
- III nickel-iron-chrome alloys
- IV nickel-chrome-iron alloys
- V nickel-chrome-cobalt alloys

Materials of group II are not hardenable by heat treatment. Groups I, III and IV comprise both non-hardenable and hardenable alloys. Materials of main groups III and IV, which can be hardened provided they have a corresponding amount of aluminium and/or titanium in conjunction with heat treatment, are called "super alloys" as do those of group V. The respective trade name of each alloy to have been offered on the market was assigned to each main groups I–V. The classification in Fig. 7.40 is not very precise; for example, not all materials classified as nimionic contain cobalt as an alloying element, yet this classification is still good

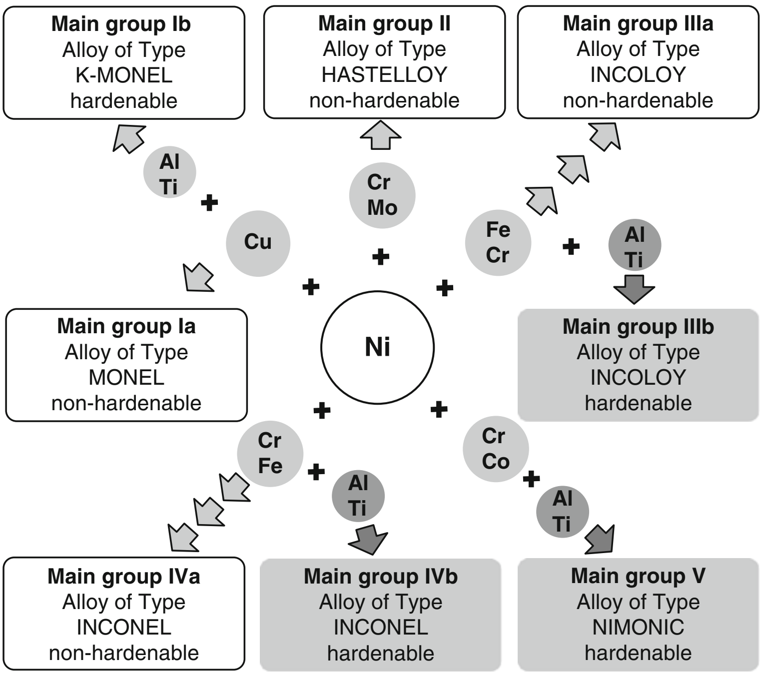


Fig. 7.40 Classification of nickel alloys in main groups (Source: Wiggin Alloys, Huntington Alloys), acc. to EVERHART [Ever71]

for orientation purposes. It should be noted that every manufacturer provides its products with its own trade name.

The main stress of this chapter is on hardenable high temperature resistant nickel-based alloys, which are chiefly utilized in aeroplane engines and stationary turbines. The alloying elements of these materials can be grouped into three categories in accordance with their effect on the microstructure. Some elements effect it in several respects.

The elements Cr, Co, Mo and W form mixed crystals with nickel. Besides an increase in strength at low temperatures, they cause an increase in the alloy's creep strength at high temperatures because dislocation creep is limited in the γ mixed crystals [Schu04]. Chrome improves oxidation and corrosion resistance, cobalt promotes the stability of the γ' phase.

The most essential strength-increasing mechanism among high temperature resistant nickel-based alloys is the deposition of intermetallic phases γ' (Ni_3(Al,Ti)) and γ'' (Ni_3(Nb,Al,Ti)). The γ' phase, which is coherent to the γ matrix, causes particle-hardening of the structural matrix that remains effective up to the high temperature range. It is formed by alloying aluminium, which can also be substituted with titanium and tantalum. The γ' phase can already originate from the molten bath. In the cast state however, it exhibits an uneven particle size, formation and distribution in the microstructure. For this reason, wrought alloys are subjected to a multistage heat treatment after melting in a vacuum and solidification. Solution annealing first dissolves the γ' phase in the mixed crystal matrix. During cooling from solution annealing temperature, it is deposited again in a – in comparison to the cast state – more consistent form in the structural matrix. Final storage at high temperatures brings about a further improvement of the consistency of the particle size and form of the γ' phase in the structure [Schu04].

In the case of materials that have an increased amount of niobium, such as Inconel 718, there is also a hardening via the γ' phase in addition to the γ'' hardening. In contrast to the γ' phase however, the γ'' phase tends toward a more rapid coarsening and thermal instability, which limits the long-term use of Inconel 718 to 650°C [Bürg06, Kenn05].

More recent nickel-based alloys, such as Allvac 718Plus, thus have a larger amount of titanium and cobalt than Iconel 718 with the same Nb-content in order to improve the formation and stabilization of the γ' phase. For example, the continuous operation temperature of Allvac 718Plus was able to be raised to 704°C. In the case of the nickel-based alloy Udimet 720, the γ' phase appears in the shape of Ni_3(Al,Ti) and $(Ni,Co)_3$(Al,Ti). A higher amount of cobalt (14.7%) reduces the solubility of titanium and aluminium and thus makes it possible for the γ' phase to form in average temperatures. Udimet 710 is applicable for continuous operation temperatures of up to 730°C [Tors97, Helm00, Mark05, Bürg06].

In the case of polycrystalline alloys it is differentiated between forging alloys and casting alloys. Forging alloys, such as Inconel 718, Waspaloy, Udimet 720LI and Allvac 718Plus contain a volume percentage of about 30–40% of γ' phase. With increasing amounts of γ' phase, deformability and machinability become increasingly limited. In the microstructures of cast alloys however, such as Inconel

713C and MAR-M-247LC, there are larger amounts (up to about 70%) of γ' phase [Schu04]. Because of the larger amount of γ' phase, cast alloys are no longer forgeable and are generally machined using grinding methods. Machining using methods with geometrically defined cutting edges is possible to some extent, depending on the cast alloy. Due to the poor machinability of cast alloys, usually fine casting methods are used close to the final contour that do not require extensive mechanical after treatment [Schu04].

The elements Cr, Ti, Mo, W and Ta form carbides of various chemical compositions (MC, M_6C and $M_{23}C_6$) with carbon, which is present in small concentration in the alloys. In polycrystalline alloys, these carbides are formed chiefly at the grain boundaries. They bring about an increase of creep resistance because they inhibit sliding of the grain boundary [Schu04].

In addition to cast-metallurgical fabrication, parts made of nickel-based alloys can also be produced using powder-metallurgical methods. The goal of PM engineering is to avoid all forms of segregation during solidification as well as concentration and structural gradients. In addition, it is possible to increase the strength of alloys by pulverizing and compacting ultrahard, non-forgeable alloys (cast alloys) generally containing high amounts of γ' phase. PM technology makes it possible to develop alloys with a higher high temperature strength than forging alloys. One central problem of PM technology is that no foreign components may infiltrate the powder, since even those of the same size as the powder particles can lead to cracking in the sense of fracture mechanics. The powder is encapsulated, hot-isostatically compressed and then forged in order to decrease the possibility of a larger foreign component inclusion or remaining inhomogeneity. One method variant in this context is “gatorizing”. This is a process in which the powder is isothermally forged using a creep forming process. PM alloys are used to manufacture turbine blades for both civil and military aeroplane engines [Adam98].

The varying chemical compositions and crystalline structures of nickel-based alloys have a direct effect on their machinability. With respect to their machinability, nickel-based alloys can be divided into five machinability-groups (Fig. 7.41). Here, group 1 represents easier, 3 average and 5 difficult levels of machinability. Representative materials are listed for each machinability-group, which are designated by their trade name as is customary for nickel-based alloys.

In the case of alloys of machinability-groups 1 and 2, cold-shaping (strain-hardening) has a positive effect on chip formation and surface quality. The machinability of these groups is comparable to that of corrosion-resistance austenitic steel.

Machinability-groups 3 and 4 comprise nickel-based alloys that are hardenable by heat treatment, also called “super alloys”. With respect to tool wear and potential surface quality, roughing of the materials of both groups should take place in a solution-annealed state and finishing in a hardened state. With respect to machining, there is practically no difference between wrought and cast alloys of groups 1 and 4 with the same composition. Due to their higher strength and usually high amount of γ' phase in the structure, PM alloys are assigned to machinability-group 4. Cast alloys of group 5 are very difficult to machine because of the large amount of γ'

Machinability group				
1	2	3	4	5
Wrought alloy				Cast alloy
Alloy of main group I.) Ni-Cu Leg.	Non-hardenable alloys of main group I.) Ni-(Cr)-Mo Leg. III.) Ni-Fe-Cr Leg. IV.) Ni-Cr-Fe Leg.	Hardenable alloys of main group II.) Ni-(Cr)-Mo Leg. III.) Ni-Fe-Cr Legierungen IV.) Ni-Cr-Fe Legierungen V.) Ni-Cr-Co Legierungen		High temperature cast alloys
Examples Monel 400 Monel 401 Monel 404 Monel R 405	Examples Hastelloy B Hastelloy X Incoloy 804 Incoloy 825 Inconel 600 Inconel 601	Examples Incoloy 901 Inconel 718 Inconel X750 Nimonic 80 Waspaloy Allvac 718Plus	Examples Nimonic 90 Rene 41 Udimet 720LI Astroloy LC PM René 95 PM	Examples IN100 Inconel 713C Inconel 718C Mar-M247LC Nimocast 739

Fig. 7.41 Classification of nickel-based alloys in machinability-groups (Source: Machining Data Handbook, Huntington Alloys)

phase and of carbides, their coarse-grain microstructure and low grain boundary strength; fractured material particles and cracks in the grain boundaries frequently cause problems when manufacturing functional surfaces [Lenk79].

Due to their mechanical, thermal and chemical properties, nickel-based alloys are generally included among materials that are hard to machine. Their high high-temperature strength in comparison to steel, low thermal conductivity, their considerable tendency to form built-up edges and to strain hardening – as well as the abrasive effect of carbides and intermetallic phases – lead to extremely high levels of mechanical and thermal stress on the cutting edge during machining. Under the machining conditions shown in Fig. 7.42, 60% higher cutting forces and about 35% higher temperatures were measured on the chip bottom side when turning Inconel 718 with $v_c = 50\,\mathrm{m/min}$ compared to the heat-treated steel 42CrMo4 + QT [Kloc06b]. Due to the high thermal and mechanical stress, tools made of high speed steel and cemented carbide can only be used at low speeds when machining nickel-based alloys. Common cutting speeds when turning Inconel 718 and Waspaloy with uncoated cemented carbides of ISO-application group HW-K10/20 are $v_c = 20{-}50\,\mathrm{m/min}$. In the case of finish turning, cutting speeds of up to 100 m/min can be reached with coated cemented carbides.

Presently, turning operations are still executed to a large extent with cemented carbides. In the case of pretreatment under average roughing conditions however, ceramic materials are being used increasingly – in the case of finishing both ceramic and PCBN cutting tool materials.

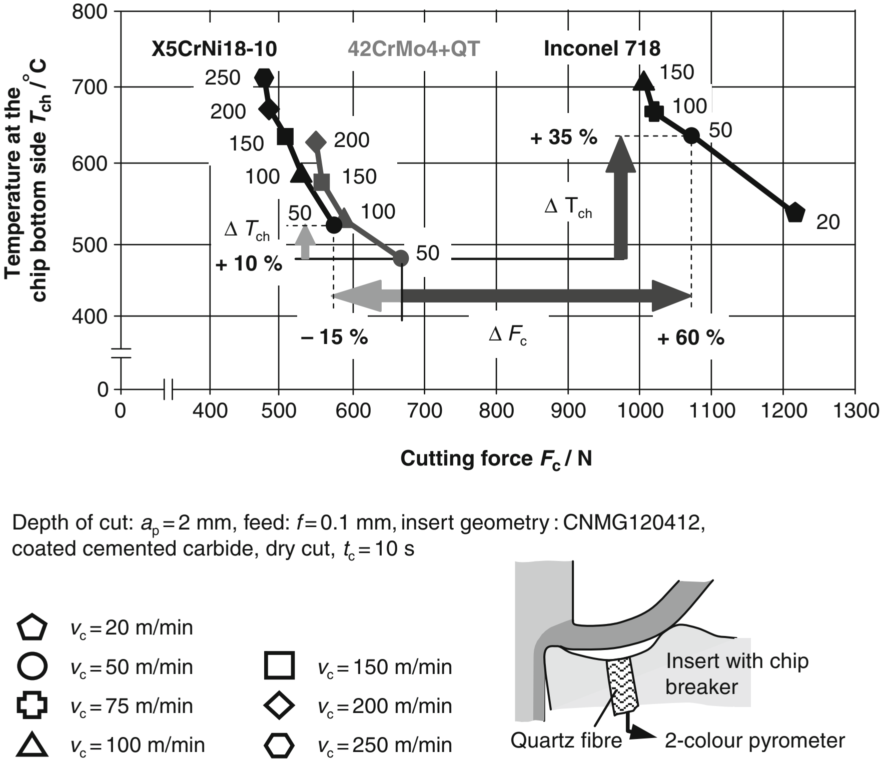


Fig. 7.42 Comparison of measured cutting forces and chip bottom side temperatures during turning of different types of steel and the nickel-based alloy Inconel 718

In many machining tasks today, the use of cutting ceramics is the state of the art. Of the ceramic cutting materials, composite ceramics ductilized with SiC whiskers (CW) have shown the greatest performance potential so far. They are suited both for finishing and for pretreatment of turbine parts under average roughing conditions ($v_c = 150-300\,\text{m/min}$, $f = 0.12-0.3\,\text{mm}$, $a_p = 0.5-2\,\text{mm}$). Of the silicon nitride cutting ceramic group (CN) especially the α/β-SiAlONs have proved equal to whisker-reinforced cutting ceramics in their wear properties ($v_c = 150-200\,\text{m/min}$, $f = 0.12-0.3\,\text{mm}$, $a_p \leq 4\,\text{mm}$). α/β-SiAlONs consist of needle-shaped β-Si_3N_4 and globular SiAlON crystals. In this way, they combine the toughness of β-Si_3N_4-ceramics with the high hardness and wear resistance of SiAlONs in one cutting material [Gers02].

CBN cutting materials are primarily used for finishing components made of nickel-based alloys (v_c = 250–350 m/min, f = 0.12–0.2 mm, $a_p \leq 0.5$ mm). When machining nickel-based alloys with CBN, the selection of a type of cutting materials suited to the particular machining task is of primary importance. CBN cutting tool materials available on the market can differ considerably with respect to the modification and amount of boron nitride, grain size and the structure of the binder phase. The resultant chemical, physical and mechanical cutting tool material properties affect the wear and performance of CBN tool to a large extent. For nickel-based alloy finishing, fine-grain CBN types with a TiC or TiN-based binder and a percentage of 50–65 vol% of cBN have proven especially suitable [Gers02].

The arc-shaped profile of its tool life graphs (Fig. 4.54) is characteristic of turning nickel-based alloys with ceramics and CBN cutting tool materials. This is the result of various wear phenomena which are predominant in dependence of the cutting speed. While flank face wear is the primary wear criterion limiting tool life in the case of turning with cemented carbide tools, tool life is limited at lower speeds by notch wear and only at higher speeds by rake and flank face wear in the case of ceramics and CBN cutting tool materials. The arc-shaped curve of the tool life graphs indicates that there is a range of optimal cutting speeds. The closer the ascending and descending branches of the tool life graphs are to each other, the more important it is to work within a range that is as narrow as possible around the tool life maximum.

Ceramic and CBN inserts with a neutral tool orthogonal rake angle ($\gamma_o = 0°$) have become standard for finishing nickel-based alloys. CBN inserts should have a small rounding of about 10 μm on the cutting edge. In contrast, ceramic cutting inserts are usually provided with a small protective chamfer to stabilize the cutting edge.

The formation of wear notches on the major and minor cutting edges of insert is characteristic of turning nickel-based alloys with cutting ceramics or CBN cutting tool materials. This significantly affects tool performance and component quality. Notch formation on the major cutting edge leads to the formation of a burr on the workpiece edge; notch formation on the minor cutting edge leads to a deterioration of surface quality (Fig. 7.43).

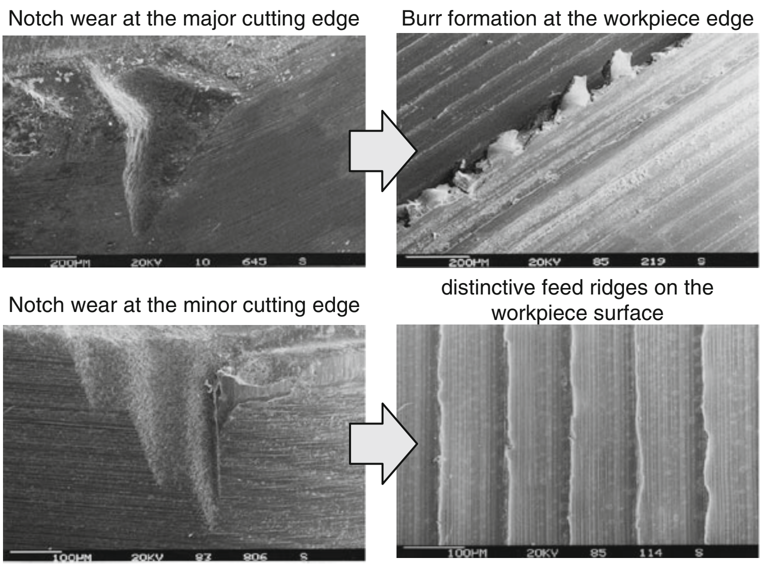


Fig. 7.43 Notch wear at the major and minor cutting edges significantly affect the performance and surface quality

Wear mechanisms that lead to the formation of wear notches include:

- fatigue, cracking and crack development phenomena – initiated by the high thermal and mechanical alternate stress on the cutting tool material as a result of lamellar chip formation and transverse material flow,
- adhesion – caused by the tendency of the kfz-crystal lattice to form microwelding and the mechanical snagging of the chip edge, flowing laterally into the notches, with the cutting tool material particles,
- abrasion – caused by the solidified, partially sawtooth-shaped chip edge, by the workpiece edge and the peaks of the feed groove pattern
- tribooxidation – caused by the chemical reaction of the workpiece material and/or the cutting tool material with components of the surrounding medium.

Wear notches arise as the result of the superimposition and mutual influence of these individual mechanisms (Fig. 7.44). A clear separation of these mechanisms with respect to their effect on total wear is only possible to a certain extent [Mütz67].

In order to reduce notch wear formation on the major cutting edge, the ratio of corner radius r_ε to depth of cut a_p should be as large as possible and the effective lead angle in the area of the contact zone end on the major cutting edge as small as possible. A small lead angle reduces the chip thickness and thus the effect of abrasion of the chip edge on the cutting edge as well (Fig. 7.45). A lead angle of $\kappa_r \leq 45°$ has proven effective for turning with tools made of cutting ceramics and CBN.

The effective use of round inserts is explicable in this context. Depending on the geometry, they exhibit not only a high amount of mechanical stability and thus

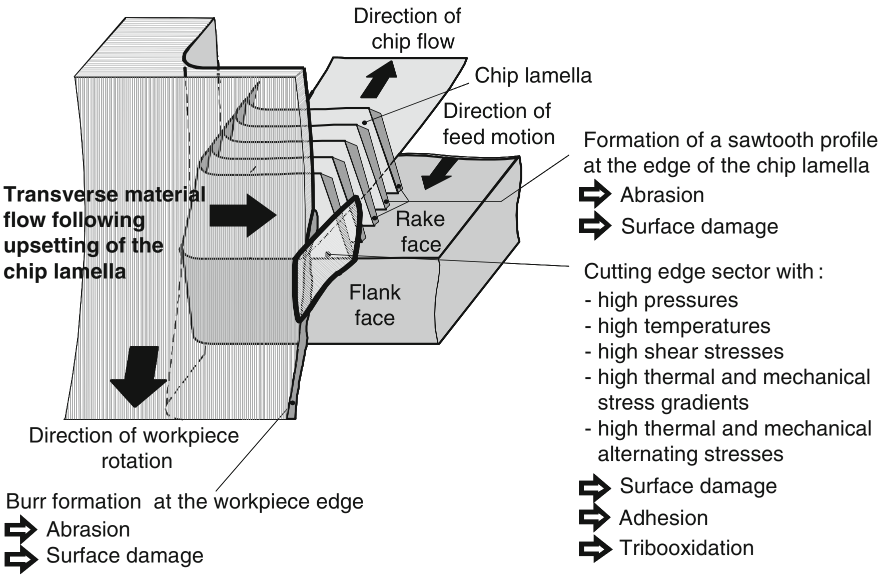


Fig. 7.44 Schematic illustration of reasons which cause notch wear at the major cutting edge [Gers98]

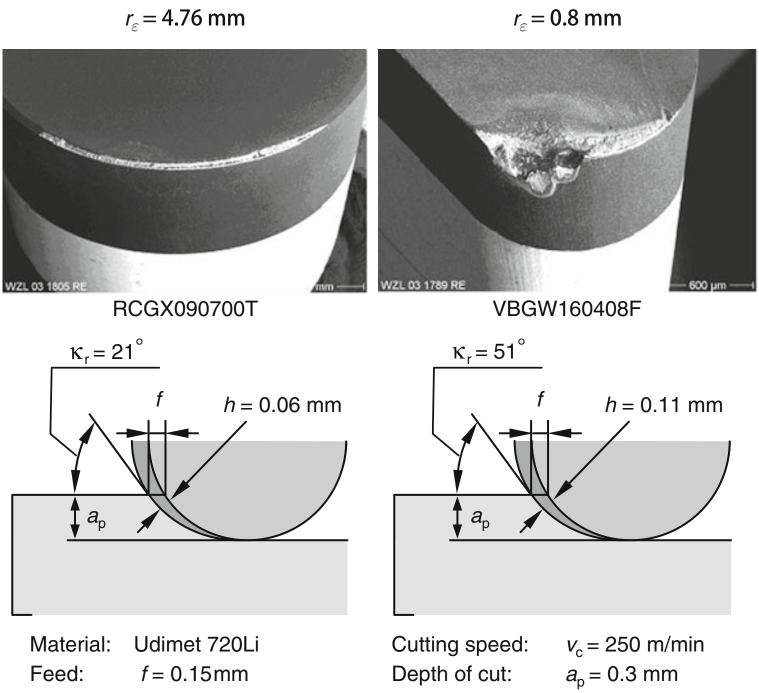


Fig. 7.45 Influence of the corner radius on wear formation during turning of a nickel-based alloy with PCBN

more resistance to insert fracture, but, especially with small depths of cut, they also generally result in small lead angles. It is thus recommendable to use round indexable inserts if the machining task allows for it. However, the disadvantage of round indexable inserts are the high passive forces, which can cause a more pronounced plastic deformation of the rim zone. Furthermore, the larger contact arc in the area of the minor cutting edge promotes the formation of wear notches. This has been observed especially in the case of turning with round indexable inserts made of cutting ceramics. If very high demands are placed on surface quality, CBN inserts with corner radii of 0.8–1.2 mm should be used [Gers02, Kloc07a].

Notch wear formation of the major cutting edge can also be effectively reduced and tool life improved by the selection of an appropriate cutting strategy. One measure that has proven very effective particularly in pre-machining with round cutting ceramics is "ramping". Here, the material is removed in pairwise pass with continuously decreasing (1st step) and increasing depths of cut (2nd step). As a result of the continuously changing depth of cut, the effect of the highly abrasive chip edge is concentrated not on one location of the cutting edge, but rather it extends along a larger area thereof. In this area, the cutting edge is indeed subject to more wear, but notch wear proceeds much more slowly, so that the insert can be used for a longer time (Fig. 7.46).

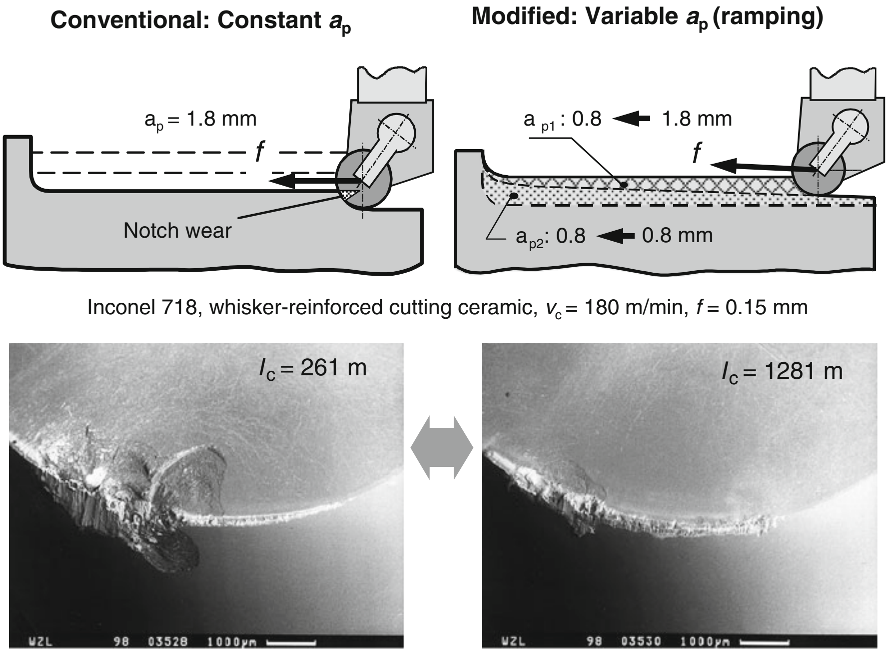


Fig. 7.46 Adapted cutting strategies allow a reduction of notch wear and a longer tool life (Source: Greenleaf)

Because of the extremely high stress on the cutting edge, milling nickel-based alloys requires cutting tool materials of high toughness and wear resistance. Because of the interrupted cut and high heat resistance of the material, the tools are subject to an extremely high mechanical and thermal alternate stress. As a result, both HSS and cemented carbide tools can only be used with relatively low cutting speeds and feeds, so that only low material removal rates can be realized. The milling of such alloys is thus an extraordinarily time-intensive machining process.

HSS tools are used with cutting speeds of $v_c = 5-10\,\mathrm{m/min}$ when machining nickel-based alloys because of their low high temperature wear resistance. Their excellent toughness properties allow however for the use of relatively large feeds ($f_z = 0.10-0.16\,\mathrm{mm}$). Due to the large number of cutting edges being engaged as well as the milling cutter with the fine knurled splines, milling with HSS tools is extraordinarily quiet. Coated end milling cutters made of PM-HSS are therefore suited above all for roughing nickel-based alloys. With a cutting edge geometry adjusted to the particular requirements of these materials, high material removal rates and a large amount of manufacturing safety is possible.

End milling cutters made of superfine and ultrafine-grain cemented carbides have a much higher level of high temperature hardness than HSS tools. They can therefore be used at much higher cutting speeds ($v_c = 20-100\,\mathrm{m/min}$), allowing for a significant reduction of machining times. They are commonly used for average roughing and especially for finishing operations. When end milling cutters and inserted-tooth cutters that are fitted with HM indexable inserts are used, the performance capacity of the tools is determined to a great extent by cutting tool material spalling on the cutting edges of the indexable inserts when nickel-based alloys are machined. They are formed as a result of a mechanical overload of the relevant section of the cutting edge. For the sake of a dynamically stable and quieter milling process, one should strive to have as many cutting edge rows as possible in simultaneous engagement. This demand cannot always be fulfilled for the given machining task due to the design of milling cutters fitted with cutting inserts [Gers02].

If grooves must be generated by rough milling in full groove cut into components made of nickel-based alloys, the tools are subject to very high levels of thermal and mechanical stress. Due to the wrap-around angle of 180° between the tool and the workpiece, a very strong force component affects the tools in the feed direction which strongly subjects the end milling cutter to bending. Faults in shape and dimension on the groove flanks are the result. Due to the high static and dynamic stress, the tools can only be used in full groove cut at relatively low speeds ($v_c = 20{-}40\,\mathrm{m/min}$) and axial depths of cut ($a_p = 0.5 \cdot D$). In the case of long-protruding tools, the cutting parameters must be further decreased in order to reduce mechanical stress and the danger of cutting edge fracture and total breakage of the tool.

Economical alternatives to conventional end milling of grooves into materials that are difficult to machine include "trochoid milling" (Fig. 7.47) and "plunge milling". In the former case, the feed motion of the milling cutter, whose diameter is smaller than the width of the groove, superimposes an approximately circular motion. Usually, down milling is used. What is special about this method is that, as a function of the ratio of milling cutter diameter to groove width and the chosen feeds in axial and radial direction, the wrap-around angle is significantly smaller than 180°, thus corresponding to finishing conditions. Trochoid milling leads to a considerable reduction of mechanical tool loading. As a result, not only the cutting speed but in particular the axial depth of cut can be significantly increased compared with conventional groove milling. Especially in the case of slender, long-protruding tools, the realizable depth of cut is several times more than that of conventional milling. Because of the larger potential depths of cut, the material volume to be machined is distributed across a larger cutting edge length. The tools are thereby used more effectively and tool life is much higher with respect to the amount of material machined, lowering tool costs drastically. A further advantage of this method is that the width of the groove that is to be manufactured is independent of the tool diameter and can be produced accurately in one milling cycle. Since in this method only low forces act upon the tools on both flanks when entering and exiting, errors in form and dimension on both groove flanks are extraordinarily few.

By superimposing a wobbling motion, not only straight but also curved flanks can be created with this method. In the case of plunge milling, the end or plunge milling cutter is plunged axially into the workpiece with low radial feed. In this way, the tool is primarily stressed in the axial direction [Kloc04]. The radial tool-bending forces are, dependent on the radial feed and the corner design of the end

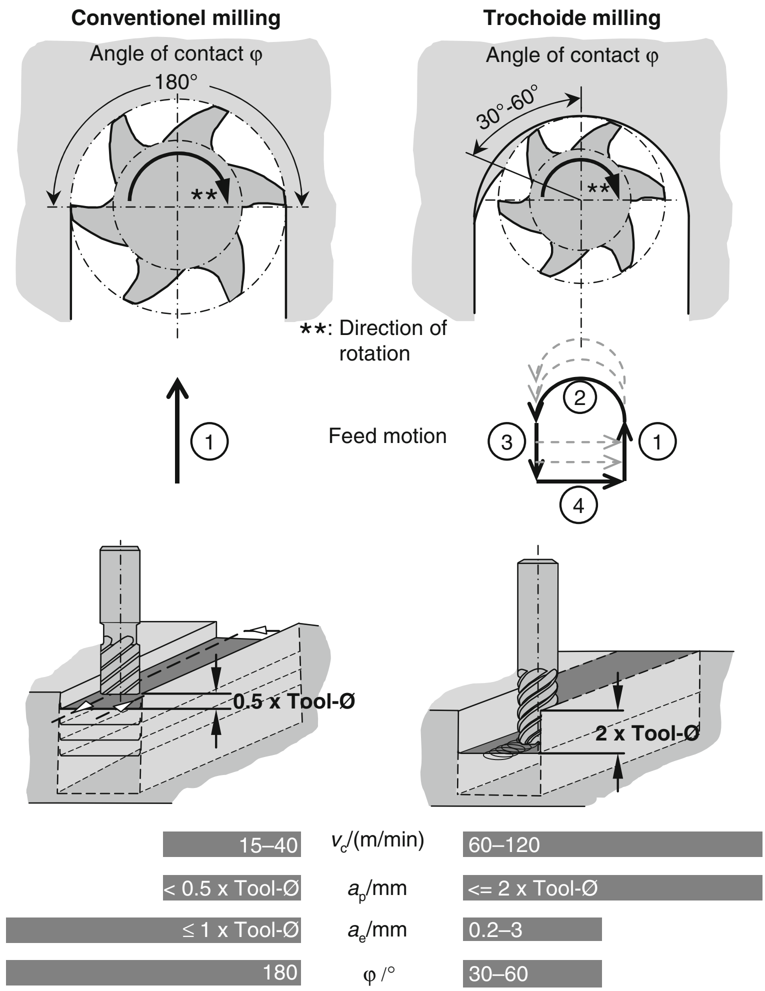


Fig. 7.47 Trochiod versus conventional end milling

milling cutter, relatively small. By plunge milling, very deep and narrow cavities can be produced. The disadvantage of this method is that the stress is concentrated exclusively on the corners of the tool.

When machining materials that are difficult to machine, the performance of an end milling cutter can be affected significantly by the micro-geometry of the cutting edge as well. In the past, popular opinion held that end milling tools should have cutting edges that are as sharp as possible to machine materials like nickel and titanium alloys because of the small realizable cross-sections of the undeformed chip. Highly positive and sharply formed cutting edges do indeed have good cutting properties, but, as a rule, the sharper a cutting edge is the greater is its jaggedness caused by cutting edge spalling. As investigations have shown, the jaggedness of the

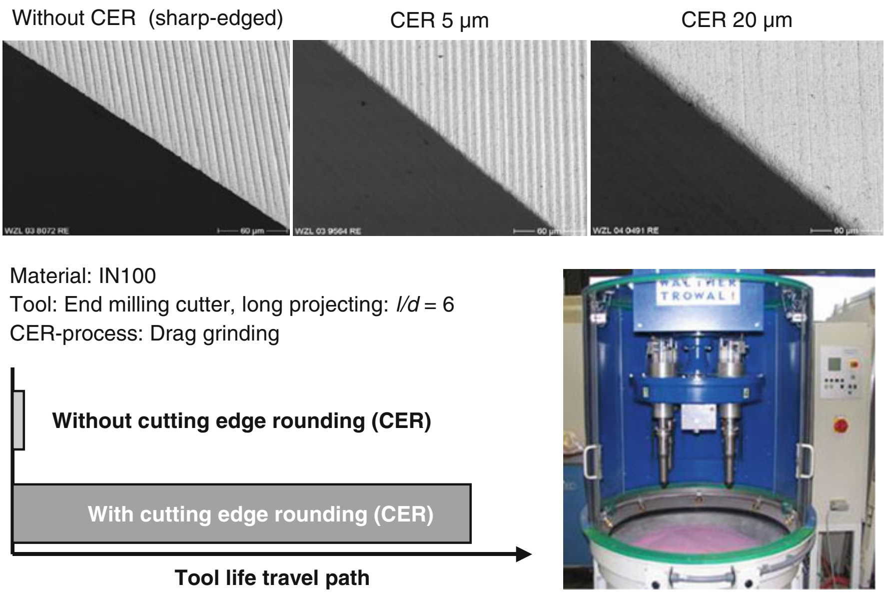


Fig. 7.48 Influence of the cutting edge rounding on tool wear during end milling of a nickel-based alloy with long cantilever length

cutting edge – produced as a function of the grinding process, cutting tool material microstructure and cutting edge geometry – influences the performance capability of a tool quite considerably. A small rounding of the cutting edge in the order of 5–20 µm can significantly increase the wear properties of end milling tools, especially when machining materials that are hard to machine (Fig. 7.48). Cutting edge rounding stabilizes the cutting edge and either evenly smoothens or removes small defects caused during manufacture by grinding in the form of micro-fractures, loosened cemented carbide crystals or micro-cracks. In this way, an even wear zone is formed on the cutting edge, and wear progresses in a continuous fashion. The cutting edge can be prepared in different ways. Besides drag finishing, used on the tools shown in Fig. 7.48, brushing, jetting with abrasive media or cutting edge rounding with laser beams can be used [Denk05].

The use of cutting ceramics based on silicon nitride and whisker-reinforced aluminium oxide when rough or finish milling with inserted-tooth cutters has opened a new dimension in the milling of Inconel 718 (Fig. 7.49). Applicable cutting speeds extend to more than 1000 m/min and are thus many time more than those commonly used for milling with cemented carbide. At a cutting speed of $v_c = 800\,\text{m/min}$ and a feed of $f_z = 0.1\,\text{mm}$ compared with cemented carbide with $v_c = 40\,\text{m/min}$ and $f_z = 0.06\,\text{mm}$, this means an increase in cutting performance by a factor of 33 [Krie02, Gers02].

The causes of such a high performance capacity are based on the chemical and mechanical stability of ceramics. However, the use of cutting ceramics still has restrictions because of their limited toughness. A sufficiently high level of toughness

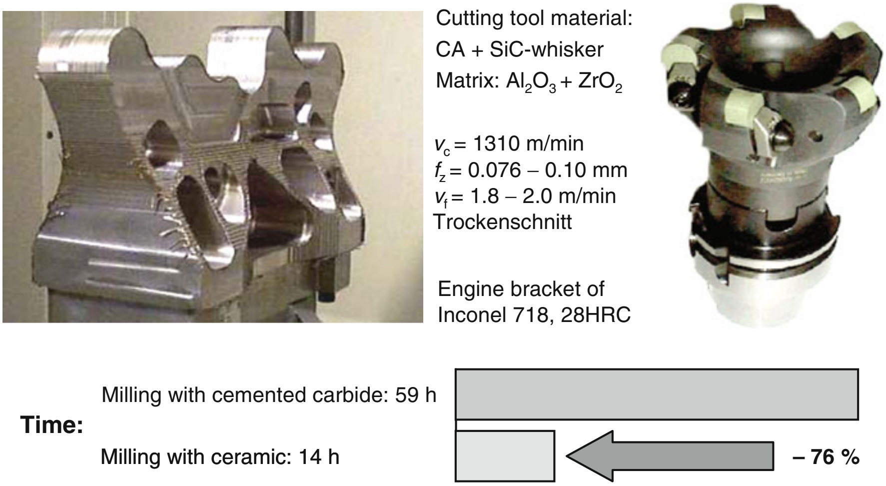


Fig. 7.49 High-performance milling of Inconel 718 with cutting ceramics (Source: Kennametal)

is necessary to avoid cracks and fractures especially in the case of interrupted cut. Reinforcing ceramic cutting tool materials with needle-shaped ß-Si_3N_4 crystals or with SiC whiskers meets this requirement and contributes considerably to the improvement of toughness. The needle-shaped crystals or whiskers embedded into the ceramic matrix can transfer tensile stress and thus reduce cracking. The energy required to loosen these crystals delays crack development. At the needle-shaped crystals, cracks extending in the matrix are forced into energy-consuming alternate routes, reducing the speed of crack development. These effects increase the toughness of the cutting tool material and its dynamic loading capacity. [Krie02].

7.7 Machinability of Non-metals

The most technically important non-metals that are commonly machined include graphite and fibre-reinforced plastics.

7.7.1 Graphite

Graphite is – like diamond – a modification of carbon and is crystallized in a hexagonal lattice structure. It is manufactured by coking coal. The fabrication process allows for a considerable degree of freedom in process design, whereby the resulting material properties can be to a great extent adjusted to the specific application. Industrial graphite is characterized by good electric and thermal conductivity and is temperature-resistant up to 3000°C. Further properties such as density, thermoshock resistance, corrosion resistance and chemical resistance can be adjusted to a large extent and across a large range by the manufacturing process. The main application areas of graphite are above all in those with high operating temperatures as electrodes, heating conductors, sealing and sliding elements as well as support materials in

- the semiconductor industry (e.g. the preparation of ultrapure silicon),
- non-ferrous metal production (e.g. heating elements in furnace installations),
- energy transfer (e.g. sliding contacts),
- machine construction (e.g. sliding rings) and
- the metalworking industry (e.g. electrodes for spark erosion).

Graphite grains are 3–15 μm large and firmly integrated in a coked binder matrix. This structural composition leads to brittle material properties with a removal mechanism which differs fundamentally from that of steel processing. This implies especially that experiences made in steel machining are difficult or impossible to relate to graphite processing with geometrically defined cutting edges. There is no plastic deformation in graphite upon penetration of the tool cutting edge into the material. Therefore, there is *no* "chip formation and removal" such as is found when machining ductile materials. Instead, the following effects predominate:

- Compressive stresses introduced under the tool cutting edge that are reduced by secondary cracking lead to material disintegration, and there is also a pronounced formation of ultrafine dust.
- Due to leading crack fronts, there is a chipping of graphite particles and formation of fracture planes in front of the tool cutting edge (Fig. 7.50).

Depending on the cutting values and engagement conditions of the tool, one of these two effects is predominant. By increasing the cross-section of undeformed chip and the cutting speeds, there is a reduced lowering of stress by secondary

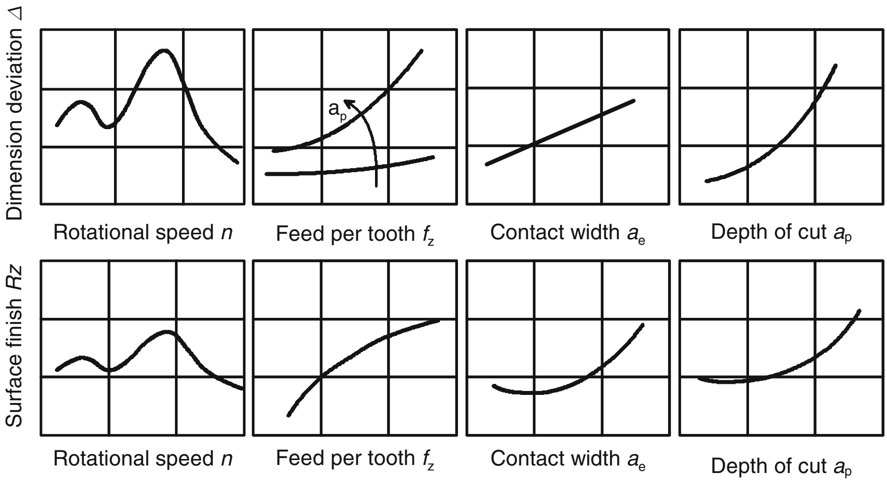


Fig. 7.50 Qualitative influences on the machining of graphite electrodes

cracks, so crack development and thus the development of larger fracture planes are favoured.

The previously described effects influence the obtainable surface quality and tool wear differently. Surface roughness increases with the grain size of the graphite and with increasing chip cross-section by enlarging the width of cut a_e, feed per tooth f_z and depth of cut a_p as well as increasing the cutting speed v_c by the formation of crack fronts leading from the cutting edge.

The dominant wear mechanism in graphite machining is abrasion due to the flow of ultrafine dust and friction (wear type: particle jet erosion). Since by increasing the cross-section of undeformed chip the formation of ultrafine dust is reduced, this also reduces tool wear considerably. In general, wear is reduced by all measures taken to remove graphite dust quickly from the machining area (blowing, dust extraction). The obvious use of wet cutting has negative effects however, since the resultant abrasive suspension increases tool and machine wear. When roughing with high material removal rates of electrodes soaked in dielectric, investigations have shown a clear increase in tool life parameters due to the reduction of ultrafine dust, which however was not reproducible in the case of finishing with small chip cross-sections.

Tool selection in graphite machining is usually made in the context of wear minimalization. Abrasion by means of hard graphite dust (particle jet abrasion) leads to craters and fractures on the rake face. Friction contact promotes wear on the flank face by the abrasive washing of the cemented carbide matrix. Improving tool life with coatings such as are familiar in steel processing (e.g. TiAlN, TiN) is only possible to a certain extent. The only effective protection against the abrasive effect of graphite dust is polycrystalline diamond (DP) as cutting tool material or the use of diamond coatings because of their high levels of hardness (Fig. 7.51).

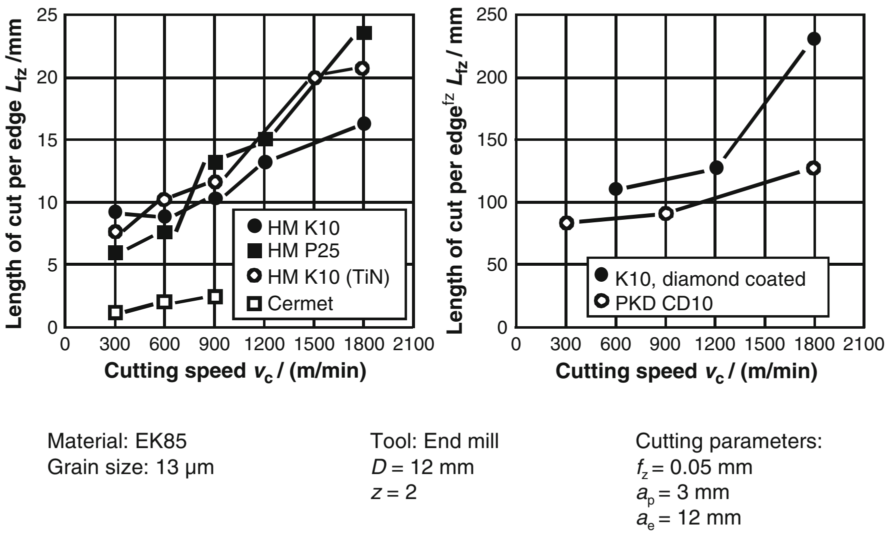


Fig. 7.51 Wear behaviour of different cutting tool materials

The rounding of the cutting edge associated with coating causes increased chipping of graphite particles and leads to a somewhat worsened surface quality. By using a positive tool inclination angle, particle jet wear is reduced due to the flatter inflow and graphite removal from the machining location is favoured. Increasing the tool orthogonal clearance angle lowers wear. Weakening of the cutting edge is of secondary importance due to the low resultant force. Since the material attributes of graphite differ fundamentally from those of metallic materials, other machining strategies can and must be employed in order to get the best results. The goal of roughing is a high material removal rate with high tool life and constant residual machining allowance. The low resultant force in graphite machining allows for a groove cut $D = a_e$) with large depths of cut. To reduce turnaround motions, an envelope-curve-limited cut distribution should be made. The goals of finishing are a high surface quality and precision of contour as well as low machining times and high tool life. As in steel machining, a distinction must be made between different machining and engagement situations. When machining curved surfaces, up milling with a falling machining direction leads, as opposed to steel machining, both to better surfaces and to higher tool lives. On steep contour areas, machining with a falling direction is preferable to a rising one, since in the former case stress on the milling cutter is in the direction of maximum tool stiffness and cutter-drift can be reduced. When machining slim webs, there is a danger of fracture on the workpiece's edges. In this case, the use of tools with a negative tool helix angle is recommendable in order to lower the risk of fracture due to tensile stresses. Up milling reduces the risk of fracture upon entry of the tool, just as down milling upon the tool's departure. In order to avoid these contrary demands, it is recommendable to work completely in an up milling process and to place the tool exit in the allowance still to be machined of a neighbouring surface. Especially when thin tools are used, material accumulation in the area of the grooves requires pre-processing along the groove before the neighbouring surfaces can be finished and the groove pulled behind. The danger of tool-drift is higher with smaller grain sizes and increasing graphite hardness.

In summary, it is clear that the material properties of graphite, with its brittle attributes, demand a fundamentally different process design in the case of milling. The low machining forces and low thermal stress make it possible to use high cutting speeds and feed velocities and thus place special demands on the tool/process kinematics/machine tool system. To reduce wear, diamond-coated cemented carbide milling cutters with cutting edge rounding and a large tool orthogonal clearance or PCD milling cutters should be utilized. The machine tool used must be capable of implementing the high potential process parameters even when machining complex geometries. To do this, high spindle rotational speeds and a high axial dynamic are required, as well as a rapid NC control. Dust formation demands efficient ventilation and filtering in an encapsulated working space. During NC programming, machining strategies that are adjusted to the peculiarities of graphite should be implemented in order to guarantee consistent component quality.

Figure 7.52 shows two graphite electrodes typically used in tool and mould construction. In the case of sinking electrodes such as are used in the manufacture of

Fig. 7.52 Examples of sinking electrodes for electro discharge machining

a computer mouse (Fig. 7.52, left) usually complex free-form geometries must be generated. Frequently, divided electrodes must be produced which are required for the fabrication of grooves in injection moulding tools.

7.7.2 Fibre-Reinforced Plastics

Fibre-reinforced plastics (FRP) consist of a duro-plastic or thermoplastic polymer matrix into which short or long fibres are embedded. Long fibres can be introduced as roving or a sheet material. To produce components made of fibre-reinforced plastics, as a rule post-production operations are required in order to form the components in accordance with their function and geometrical characteristics. As opposed to metallic materials, with which the component is often fabricated "from solid", it is attempted to fabricate components made of FRP close to the final contour so that the ratio of the volume machined to the total volume remains small.

Fibre-reinforced plastic press parts can be deburred both by methods with geometrically defined and geometrically undefined cutting edges. Since this volume of the compendium deals exclusively with machining methods with geometrically defined cutting edges, these methods will be in the foreground. Of this subgroup of manufacturing processes, milling and sawing have proven the most effective. If functional surfaces must be manufactured from plastic, high demands with respect to formal and dimensional accuracy or high levels of surface quality cannot be met, or at least only at unwarrantable cost. For example, only grinding methods are used as the typical finishing method for producing clearances such as bearing seats [Wuer00, Köni90b, Wien87, Wuns88].

To machine a component made of fibre-reinforced plastic, the mechanical and thermal material properties are of especial importance. These are affected during machining essentially by the properties of the fibre type used in the plastic.

This has a decisive influence on method selection or also, for example, the suitability of tool designs. The fibre is characterized by high tensile strength, its elastic modulus and this low fracture strain compared to the matrix. In addition, it has, depending on the fibre type, highly varied thermal parameters, which sometimes deviate a great deal from those of the plastic matrix. However, reinforcement fibres have differing properties under mechanical strain in accordance with their respective structure. Figure 7.53 clarifies this by looking at the fracture behaviour of single fibres in different load cases.

Certain conclusions can be drawn from the results of the shear experiments, with or without axial prestressing, regarding the “machinability” of the respective fibre types. Glass and carbon fibres exhibit brittle fracture properties under tensile, shear or bending strain, whereby the fracture plane of carbon fibres is somewhat rougher. The much tougher aramid fibres tend on the other hand to deflect the cutting edge under shear or bending stress, which can lead to a “frayed” cut surface during the machining process. These fibres can only be satisfactorily cut when prestressed and are frequently divided axially [Böns92].

The properties of the matrix have more of an effect on process management. Its “machinability” is characterized by the low elastic modulus, low strength, high fracture strain and above all by low temperature resistance of plastics. As opposed to thermosetting plastics, which stay in a solid state until their decomposition temperature is reached, the thermoplastics used for composite materials soften beyond a temperature of about 200°C. This further lowers the already low temperature resilience of the matrix. The low temperature conductivity of the polymer matrix materials is a further problem. As a result of this, the process heat introduced to the material can only be poorly removed from the active site, and thus the component is damaged by burns and the tool by thermal strain (Fig. 7.54).

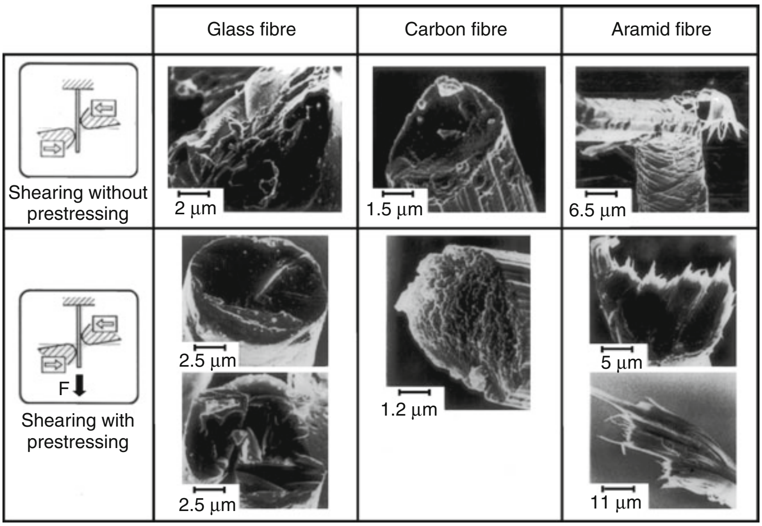


Fig. 7.53 Fracture behaviour of different fibre types

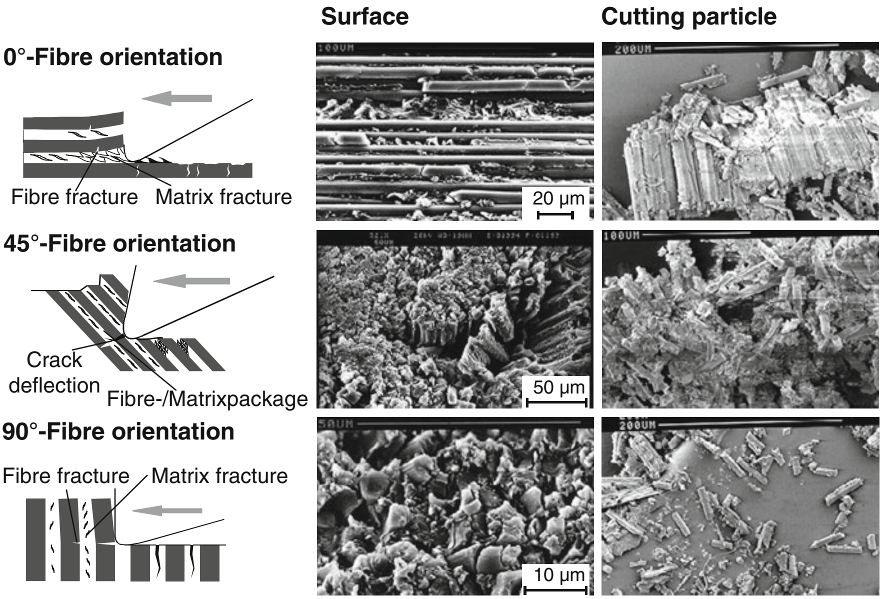


Fig. 7.54 Failure mechanisms during machining of fibre-reinforced plastics

Due to its inhomogeneous structure, the "machining" of fibre-reinforced plastics involves the formation of highly diverse "chips" compared with metallic materials. The development of very fine particles made of fragments of the base material and of fibres, causing a large amount of dust, is characteristic of machining fibre-reinforced plastics [Rumm96, Wuer00]. The quality of the machined surface is dependent to a large extent on the orientation of the fibres in the polymer matrix.

When machining under the 0°-fibre orientation, the particles formed are primarily fibre and matrix packets, which lead to a high level of machined surface quality. If on the other hand it is machined with a 90°-fibre orientation, particle formation of single fibres and matrix fragments is predominant. The rim zone of the machined area near the surface is then characterized by microcracks. Machining with a 45° or 135°-fibre orientation leads to a superimposition of the above particle formation types. Due to the high levels of mechanical stress, the surface is always heavily damaged by the cutting edge. This results in high surface roughness values as well as the formation of very small particles made of fibre and matrix packets.

Measurements of these particles have shown that respirable sizes do exist (particle $< 5\ \mu m$) that can have a negative effect on the health of the operator. An efficient suction and filtering of these particles is therefore absolutely necessary. In addition, it is advisable to seal off the working area.

The most common machining method when manufacturing components made of fibre-reinforced plastics are turning, milling and drilling [Köni91, Tras92]. Selection of the most proper tool should be made in analogy to machining metallic materials with reference to the technological, economical and ecological limitations.

7.7.2.1 Machining Method: Milling

During machining, the high hardness of glass fibres and in particular of carbon fibres leads to pronounced wear phenomena on the tools employed.

Such tools must therefore have a high level of resilience against abrasion and of toughness. Suitable cutting tool materials are ultrafine-grain cemented carbides (HF), polycrystalline diamonds (DP) or diamond-coated cemented carbide tools (HC) of application group K10.

To make a precise separation of the fibres possible, the cutting edge must be very sharp. With respect to the geometry of the cutting edge, it must have a very low level of raggedness, and the cutting edge radius should be in the range of the fibre diameter (~10 μm). The grinding of a chamfer on the flank face of the milling tool has also proved effective. In this way, elastic rebound of the fibres embedded in the base material can be considerably reduced, reducing thermal and mechanical stress during machining. A classification into areas of application is made according to the type of fibres, their length and their percentage in the composite material. Untwisted two-edged milling cutters with sharp cutting edges should be used to mill components made of directed long fibres with a large fibre amount, since only such a tool can separate the fibres cleanly. DP-fitted milling tools are superior to cemented carbide milling cutters with respect to realizable tool life parameters and surface quality, but they involve much higher acquisition costs.

Because of the large number of possible material combinations, generalizations can be made regarding optimal cutting parameters. In many cases, cutting speeds of 800–1200 m/min with moderate feeds per tooth have yielded successful machining results.

For a high-quality machining of aramid fibre-reinforced laminate, standard cutting edge geometries can be used so long as it is taken into consideration that the fibres can only be acceptably cut when prestressed. This prestressing can be achieved with special tools that have been specially designed for this purpose (Fig. 7.56).

Further tool requirements include a high level of cutting edge sharpness combined with a small cutting edge radius and a high level of rake and flank face surface quality in order to minimize friction on the workpiece [Köni90b]. As in the case of glass fibre and carbon fibre-reinforced plastics, the use of tools made of cemented carbides of application group K10 have also proven effective here.

Figure 7.56 shows a distinction made on the basis of the thickness of the component to be machined. For thin workpieces, tools twisted in two ways towards the middle were developed that make it possible to change the distribution of the machining forces acting on the workpiece. To use these tools successfully, the tool must be precisely aligned with the workpiece. For thicker components, it is advisable to use a two-way mill twisted in contrary directions. In the case of this tool

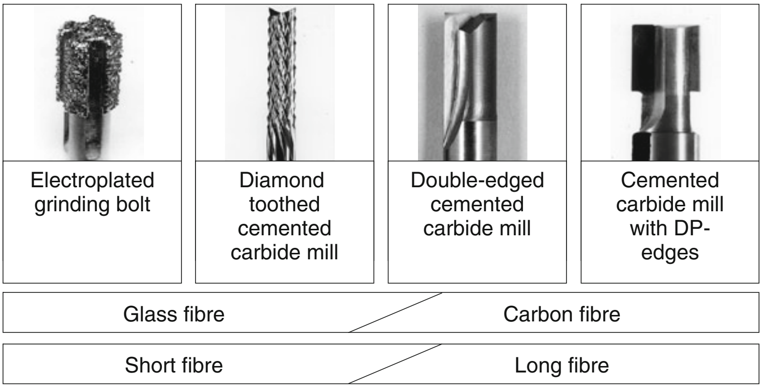


Fig. 7.55 Tools for machining fibre-reinforced plastics (GRP, CRP)

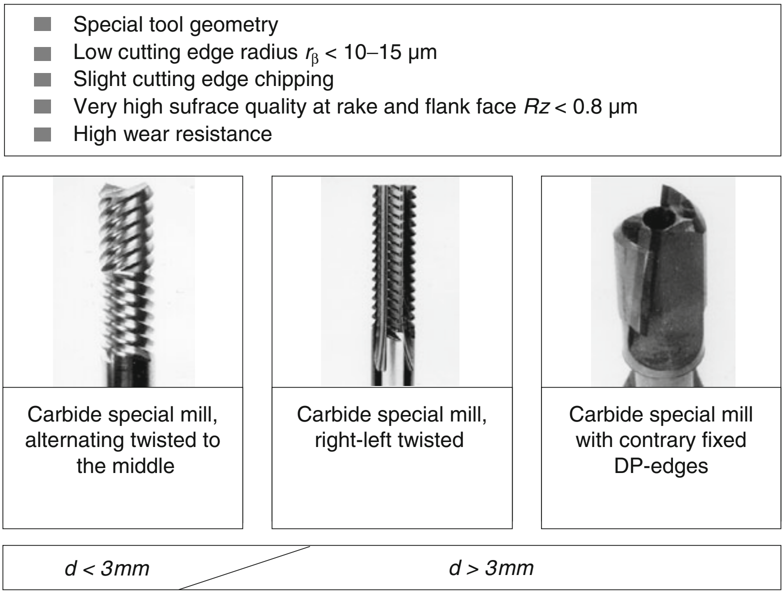


Fig. 7.56 Tools for machining aramid-fibre-reinforced plastics

variation as well, the permanent alternate stress prevents the fibres from eluding the cutting edge.

The following describes difficulties involved with particles arising when machining fibre-reinforced plastics (dust, fibre particles).

Figure 7.57 clarifies the high concentration of particles that remain in the working space even after the machining process and that is only gradually reduced. Especially the small particle sizes, thoracic (4.5–10 μm) and alveolar (< 4.5 μm) dusts, remain in the air long after machining is finished and sediment very slowly. As

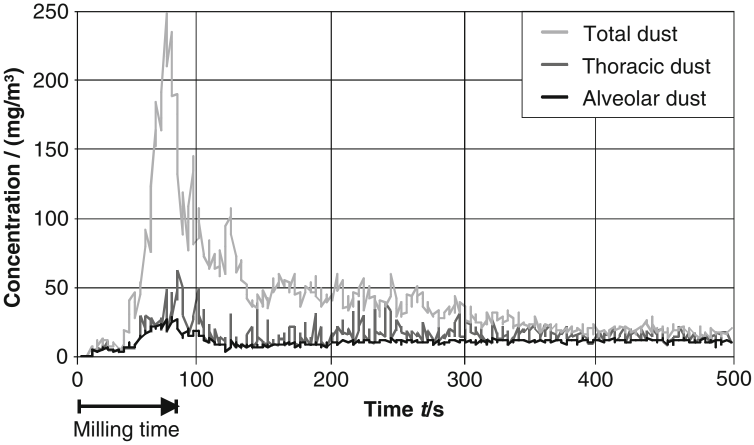


Fig. 7.57 Dust particle size during milling of fibre-reinforced plastics

an experiment, a 90 s long machining process was conducted and the machine was immediately switched off. It is important to know that even after 500 s, the legally prescribed fine dust limit of 1.5 mg/m^3 was not yet reached. These particle sizes are highly carcinogenic, so this limit must be strictly observed. The percentages of these sizes in the entire volume of dust can be influenced by tool selection and the cutting parameters. The experiment shows that it is nonetheless absolutely necessary to isolate the working area and to suction off and filter the resulting particles.

7.7.2.2 Machining Method: Drilling

In the case of drilling, it is necessary just as in milling to differentiate in accordance with the materials to be machined, glass-fibre and carbon fibre-reinforced plastics or aramid fibre-reinforced plastics. To drill CFP and GFP, highly wear-resistant tools with standard cutting edge geometries can be used as long as the cutting edge radius is as small as possible. In many cases, solid cemented carbide tools produce good results. Tools that are equipped with DP cutting edges are far superior, but they can only be used for manual drilling operations to a limited extent.

Standard tools cannot be utilized to machine aramid fibre-reinforced plastics. In this case, special cutting geometries should be used in order to guarantee separation of the fibres while prestressed (Fig. 7.58). Cemented carbide drills with very positive rake angles have become established for this. Besides a small cutting edge radius, high levels of surface quality on the flank and rake faces are required so that friction between the chip and the tool and between the workpiece and the tool reduces potential adhesion of the material on the cutting edge.

Potential surface roughness depends heavily on the type and length of the fibres. When the tool is rotated, all fibre lengths are always cut, so that, depending on the orientation of the fibres, there is a highly aberrant surface formation in the drill hole.

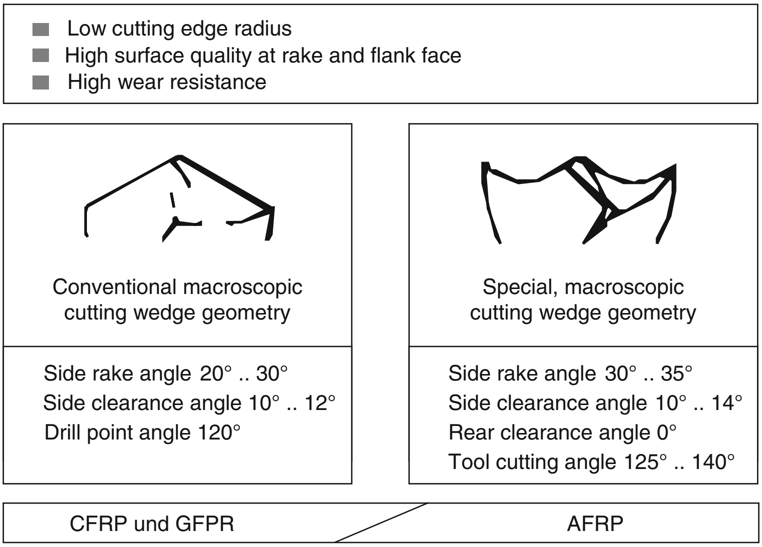


Fig. 7.58 Tools for drilling fibre-reinforced plastics

It is characteristic of the material that, in the case of fibre-orientations of 105–135° with a simultaneous compressive load, only very poor surface qualities can be realized. As opposed to CFP and GFP, there are no fibre fractures in aramid fibre-reinforced plastics, but the fibres are pulled far out of the cut surface and bent against the cut surface. This is due to the tough material properties and poorer adhesion of the aramid fibres to the base material.