

Lecture Notes in Production Engineering

Frank Vollertsen *Editor*

Micro Metal Forming

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Preface

Currently, our life is characterized by a continuous densification in all aspects and parts. Communication is faster thanks to smartphones, enabling one-to-one video, and audio contact nearly at any time and location. Smartphones have a large number of functions, which were realized before by nearly a dozen of aggregates. Comfort and safety in cars is increased by a steadily growing number of safety and assist systems, controlling not only the status of the car functions itself, but also the agility of the driver and the actions of the traffic around the car. Also, health care components comprise numerous functions in aggregates of reduced size. All those developments are the drivers for miniaturization of parts and components which have to be produced in large lot sizes at low costs. As metals are still excellent multi-functional materials, the methods of metal forming could be a means of covering these demands, if those processes could be used for the production of miniature components.

A first hot spot of micro metal forming in Germany was developed by Prof. Engel and Geiger, starting in the 1990s in Erlangen. Discussion of the scientific advances, achieved also in other countries like Japan and USA was held (not limited to but also) in The International Academy for Production Engineering (CIRP), documenting the milestones in numerous papers and two keynote papers. These keynote papers ('Microforming', issued in 2001, and 'Size effects in manufacturing of metallic components', issued in 2009) are key documents on the development of micro metal forming. The relevance of size effects is due to the fact that those effects are the reason why knowledge from (macro) metal forming cannot be transferred easily to the micro range. The effects have been the topic of a priority program funded by the Deutsche Forschungsgemeinschaft (DFG) in the period from 2002 to 2008. In 2007, a Collaborative Research Center (CRC, SFB 747) was started in Bremen, funded by the DFG and the University Bremen. Within this CRC are about 50 scientists' work on topics relevant for the further development of micro metal forming. Taking advantage of the structure of a University, lectures on micro metal forming are introduced and held continuously by the CRC-scientists. The knowledge gathered and produced in the CRC is condensed in the hand book. It is compiled to help those people who start to work

in this special field of micro metal forming. It is not only for students, doing their Bachelor or Master thesis, but also for engineers working in research and development in institutes and industry. Basic knowledge of physics, mathematics, and the principles of metal forming are required to use this book.

The aim of this book is to present the most important basic knowledge, the current state-of-the-art, and research results useful for future application of micro metal forming. Hence, the book is composed from three blocks, each comprising some chapters:

The first block (Chaps. 1–3) describes the basics in metal forming like tribology, formability, and mechanical strength. A special emphasis is laid on the questions which arise when reducing the size to the micro range. Therefore, size effects are also described which should help to understand the principle differences in micro forming besides the small size and weight of the workpieces.

Information on processes of bulk and sheet forming is given in the second block (Chaps. 4–6). Besides the effects in micro metal forming using the common processes, special processes which make use of size effects are explained in Chap. 6. The processes are only feasible in micro processing and demonstrate the positive nature of size effects. The facts explained in the first block are helpful to understand the nature and limits of those methods.

Tools in the broader sense, which are used for micro metal forming, are addressed in the third block. Not only the forming tools, e.g., the components which transfer forces on the workpiece during the forming process are meant, but also means for their production and other aggregates like forming machines, methods for quality assurance, and simulation tools. These topics are described in Chaps. 7–10.

All the authors of the 10 chapters and sections are working in the relevant fields of the CRC. They are directors or staff members of the following institutes, listed in alphabetical order together with the most important research areas covered within the CRC. The names of the directors are given in brackets:

BIAS—Bremer Institut für angewandte Strahltechnik: Laser material processing, sheet and bulk metal micro forming (Prof. Dr.-Ing. Frank Vollertsen); optical metrology (Prof. Dr. rer. nat. Ralf Bergmann)

BIBA—Bremer Institut für Produktion und Logistik: Logistics and simultaneous engineering (Prof. Dr.-Ing. Bernd Scholz-Reiter)

BIMAQ—Bremer Institut für Messtechnik, Automatisierung Qualitätswissenschaft: Process control including metrology, quality assurance (Prof. Dr.-Ing. Gert Goch)

bime—Bremer Institut für Strukturmechanik und Produktionsanlagen: Bulk metal forming including machine development (Prof. Dr.-Ing. Bernd Kuhfuß; process chain layout and automatization (Prof. Dr.-Ing. Kirsten Tracht)

IFS—Institut für Statistik: Monte Carlo-simulation and statistics (Prof. Dr. Mag. rer. nat. Werner Brannath)

IWT—Stiftung Institut für Werkstofftechnik: Physical vapour deposition, heat treatment, and mechanical testing (Prof. Dr.-Ing. Hans-Werner Zoch)

LFM—Laboratory for Precision Machining: Cutting, machining, and polishing
(Prof. Dr.-Ing. Ekkard Brinksmeier)

ZeTeM—Zentrum für Technomathematik: Industrial mathematics (Prof. Dr. Peter
Maaß), simulation systems (Prof. Dr. rer. nat. Alfred Schmidt)

The interdisciplinary cooperation between production engineering, mathematics, and physics is an excellent basis for research in the demanding field of micro metal forming. A prerequisite for successful cooperation is the funding of manpower and equipment, which was granted by Deutsche Forschungsgemeinschaft (SFB 747 Mikrokaltumformen) and University Bremen. The authors gratefully acknowledge this support. Further thanks are expressed to Dr. Sybille Friedrich for coordinating the work within the CRC including the preparation of this book.

Bremen, October 2012

Prof. Dr.-Ing. F. Vollertsen

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Global Variables

Latin

Variable	Explanation
A	Area (mm ²)
a	Distance (mm)
b	Width (mm)
d	Diameter (mm)
E	Modulus of elasticity (GPa)
F	Force (N)
h	Height (mm)
HK	Knoop hardness
HM	Martens hardness (N/mm ²)
k _f	Flow stress (MPa)
l	Length (mm)
p	Pressure (N/mm ²)
r	Radius (mm)
R	Roughness (μm)
s	Sheet thickness/layer thickness/material thickness (μm)
S	Surface roughness (nm)
t	Time (s)
T	Temperature (K)
u	Upset ratio
v	Velocity (mm/s)
V	Volume (m ³)
$\dot{V}$	Volume material removal rate (m ³ /s)

Greek

Variable	Explanation
δ	Deflection (μm)
Δ	Difference (μm)
ε	Strain ($=\Delta l/l_o$)
$\dot{\varepsilon}$	Strain rate (s^{-1})
ρ	Density (kg/m^3)
φ	Logarithmic degree of deformation ($=\ln(l/l_o)$)
μ	Friction coefficient
λ	Wavelength (m)
ν	Kinematic viscosity (m^2/s)
ρ	Density (kg/m^3)
σ	Normal stress (MPa)
σ_M	Ultimate tensile strength (MPa)
σ_N	Nominal stress (MPa)
σ_{pl}	Yield strength/elastic limit (MPa)
$\sigma_{pl\ 0.2}$	Yield strength with 0.2 % plastic deformation (N/mm^2)
τ	Shear stress (MPa)
ψ	Potential (V)

Global list of subscripts (These subscripts are used globally)

Subscript	Explanation
0	Initial condition
1,2	Condition 1, 2
a	Arithmetic
ax	Axial
BLH	Blank holder
e	Elastic
f	Yield/feed
fr	Friction
F	(de)formation
G	Grain/gravity
la	Laser
max	Maximal
min	Minimal
M	Melting/tensile strength
n	Normal
N	Rated...
pl	Plastic
P	Punch
r	Radius, radial
t	Tangential

Chapter 1

Introduction

Frank Vollertsen

Micro metal forming is a part of the wide field of metal forming. This class of production technology is characterized by:

1. The shape of the workpiece is obtained by the plastic deformation of (predominantly) metallic materials.
2. In most cases the forces on the workpiece are induced by the coordinated action of at least two tool segments.
3. The tools may contain 0–100 % of the desired shape as analogous memory. If the shape is not stored in the tools, the relative motion of the tools has to be controlled to derive a kinematic generation of the desired shape.
4. By definition, there is no material loss in metal forming. In practice however, the preparation of the raw part and finalization of the shape of a workpiece makes some cutting operations necessary, typically incurring material loss of 20–50 %.

The specific aspects of micro metal forming may be seen from a definition by Geiger et al. [1]. In this definition micro metal forming is *the production of parts and structures with at least two dimensions in the sub-millimeter range* by metal forming. Processes for semi-finished products like wire drawing or rolling of foils are not part of micro metal forming (mmf) by this definition, while those processes might entail some of the typical challenges of mmf. To illustrate this definition, it is said that a part made by mmf should fit with a strictly linear motion through a circular hole of diameter of 1 mm. It is obvious that this definition is just a restriction in the size of the parts. Therefore, one may wonder why the well-known processes of (macro) metal forming could not be applied to mmf by just reducing the billet, blank and tool sizes. Specific features of mmf and related products are:

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1. The size or at least one of the dimensions of the produced parts is comparable with the grain size of the material used, resulting in hard to control material behavior.
2. The very small volume of the parts changes the failure behavior due to different probability of the occurrence of a defect in a particular workpiece, if homogeneous defects with low density exist in the raw material.
3. The very low weight (typically between 100 μg and 10 mg) of the (raw) parts make handling difficult due to, for example, adhesion effects. As precise handling is necessary for joining operations, joining is less desirable. Therefore parts should integrate multiple functions to reduce the number of components in an assembly and to minimize the number of handling and joining operations. This results in turn in a more complex design of at least the intermediate stages of the product. As an example, interconnected multiple parts are manufactured as one workpiece, which are separated after assembly to achieve independent electrical functions. On the other hand, the small weight might allow the use of more expensive materials. This is also true for micro forming tools, which may be manufactured from a single crystalline diamond.
4. Quality assurance becomes more difficult compared to macro parts, as many methods usually employed cannot be used for the measurement of micro part dimensions. Also the (scaled down) tolerances interfere greatly with the precision of the metrology, making the use of methods like statistical process control (SPC) impossible.

As the parts are smaller than a grain of rice, there is reference to silicon technology and the question arises whether this technology can deliver the same parts as micro metal forming. A closer look at silicon technology shows differences in at least the following four features:

1. Mmf parts are often larger than those of silicon technology.
2. The variety of materials is much greater in mmf, as many metals can be processed.
3. Parts produced by silicon technology are almost always $2\frac{1}{2} D$, e.g. they have a constant cross-section along a (limited) height (so-called prismatic structures).
4. Mmf may also be economically successful for small lot sizes.

All in all, mmf and silicon technologies should be understood as being complementary rather than competing technologies.

Mmf is already in use for the production of a few parts for the electronic industry, such as shafts for micro motors or end-caps for SMD-resistors, see Fig. 1.1. The processes which are in use for the production of these parts demanded great efforts for their (mostly empiric) development, making the processes only economically viable for very large lot sizes of some millions or even billions of parts per year. A wider use of the mmf technology class could be achieved, if the production was better planned, processes were more stable and sound control was implemented. This book is intended to be a kind of technological guidance for mmf. There is continuous growth in mmf knowledge by the

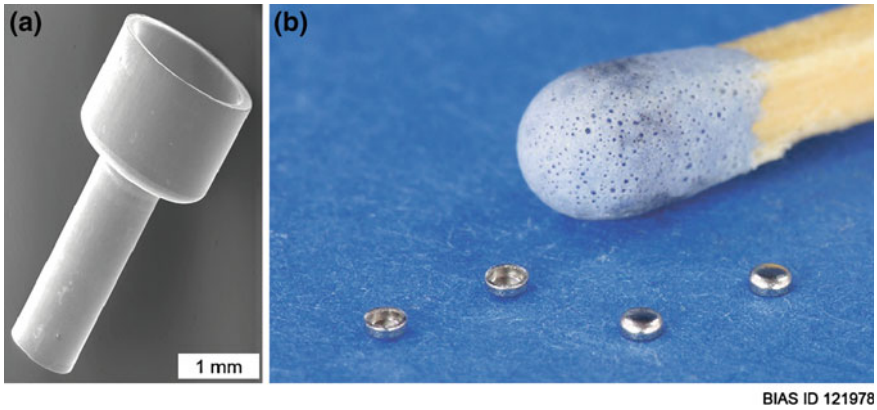


Fig. 1.1 **a** Deep-drawn shaft of a micro motor (18 drawing steps), courtesy of U. Engel. **b** Deep-drawn end caps for resistors, courtesy of Stüken GmbH & Co

further development of parts and processes. Related processes are macro metal forming, silicon technology and high-precision cutting technology, which are already described in detail. This current description should help both the new and the experienced user of mmf decide on the part and process layout for the prompt start of a stable production process.

Reference

1. Geiger, M., Kleiner, M., Eckstein, R., Tiesler, N., Engel, U.: Microforming CIRP Ann. **50**(2), 445–462 (2001)

Chapter 2

Basic Aspects

Frank Vollertsen

Latin

A	Area (mm^2)
A_0	Initial area (mm^2)
A_l	Liquid contact area (mm^2)
A_s	Solid contact area (mm^2)
a	Distance (mm)
b	Width (mm)
d	Diameter (mm)
d_0	Initial diameter (mm)
d_1	Major diameter after deformation (mm)
d_2	Minor diameter after deformation (mm)
d_p	Punch diameter (mm)
E	Modulus of elasticity (GPa)
F	Force (N)
F_{BLH}	Blank holder force (N)
F_E	Electrostatic force (N)
F_{fr}	Frictional force (N)
F_G	Gravitation (N)
F_n	Normal force (N)
F_{vW}	Van der Waal's force (N)
F_γ	Force induced by surface tension (N)
h	Height (mm)
h_0	Initial height (mm)
k	Yield stress in shear (MPa)
k_f	Flow stress (MPa)
k_{fs}	Flow stress at the surface (MPa)
k_{fv}	Flow stress in the volume (MPa)

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l	Length (mm)
l_0	Original length (mm)
l_1	Length during deformation (mm)
l_G	Grain size (mm)
l_{Na}	Length nakajima test
m	Friction factor
p	Pressure (N/mm ²)
p_l	Closed pocket pressure (N/mm ²)
q	Wear rate (m ³ · (Nm) ⁻¹)
s_0	Initial material thickness (mm)
T	Temperature (K)
t	Time (s)
v	Velocity (mm/s)
V	Volume (m ³)
w	Path (mm)

Greek

Δ	Difference (μm)
ε	Strain
φ	Logarithmic degree of deformation = $\ln(l/l_0)$
$\dot{\varphi}$	Deformation velocity (s ⁻¹)
φ_1	Major strain
φ_2	Minor strain
Λ	Scaling factor
η	Dynamic viscosity (N s/m ²)
κ	Grain number
μ	Friction coefficient
ν	Kinematic viscosity (m ² /s)
σ	Normal stress (MPa)
σ_M	Ultimate tensile strength (MPa)
σ_{pl}	Yield strength (MPa)
τ	Frictional shear stress (MPa)
τ_{SC}	Shear stress for a single crystal (MPa)

2.1 Size Effects

Frank Vollertsen

2.1.1 Definition and Categories

The term ‘size effect’ is often used when experiments with samples of different size, e.g. different sheet thicknesses in tensile tests, present non-uniform behavior. More than once in the past different behavior has been wrongly referred to as size effect. In one example, tensile stress in foils with thicknesses of 200 and 50 μm was called a size effect, despite the fact that the authors of those papers were aware of the fact that the 50 μm foil showed a stronger strain hardening than the 200 μm foil. This in turn was due to rolling without intermediate annealing. Therefore, one should be aware that not every difference observed in samples of different sizes is a size effect.

A general description of size effects as understood in the scientific context is strongly linked to the theory of similarity. This theory (see e.g. [38] for a detailed explanation) has the central hypothesis that systems will behave similarly if scaling, i.e. changing the size, is carried out according to the rules of similarity. All relevant features like speed, weight, size and so forth must be changed in a fixed relation, e.g. by a constant factor which is called scaling factor Λ . As an example, let us consider a simple upsetting test with the plastic compression of a cylinder of initial height h_0 to the final height h . In order to keep the example simple no friction should occur. The velocity v

$$v = \Delta h / \Delta t = (h_0 - h) / \Delta t \quad (2.1)$$

is the punch velocity and

$$\dot{\varphi} = d\varphi / dt = \Delta\varphi / \Delta t \quad (2.2)$$

is the deformation velocity, where φ is the logarithmic strain

$$\varphi = \ln(h/h_0) \quad (2.3)$$

In the case that h_0 is changed from size 1 to size 2, a similar material behavior (in terms of flow stress k_f) is expected for similar deformation speeds (and identical material, i.e. grain size, predeformation, temperature and so on), e.g.

$$k_{f1} = k_{f2} \quad (2.4)$$

for

$$\dot{\varphi}_1 = \dot{\varphi}_2 \quad (2.5)$$

For identical strain Eq. (2.5) yields when using Eq. (2.2)

$$\Delta t_1 = \Delta t_2 \quad (2.6)$$

As Δh_1 is not equal to Δh_2 , v_1 is not equal to v_2 . Defining

$$h_{0,1} = \Lambda h_{0,2} \quad (2.7)$$

the relation of punch velocities from Eq. (2.1)

$$v_1/v_2 = (h_{0,1} - h_1)/\Delta t_1 / (h_{0,2} - h_2)/\Delta t_2 \quad (2.8)$$

will together with Eqs. (2.6) and (2.7) yield

$$v_1/v_2 = (\Lambda h_{0,2} - \Lambda h_2)/(h_{0,2} - h_2) = \Lambda \quad (2.9)$$

Therefore, if the height $h_{0,1}$ is $\Lambda = 10$ times the height $h_{0,2}$, the punch velocity has also to be increased by a factor of 10 to obtain similar material behavior—at least according to the theory of similarity. In fact this identical material behavior is not observed in small samples, where the grain size is in the order of the sample diameter. Such deviations from the expected behavior are called size effects, which are visible under conditions where the rules of similarity are obeyed.

Size effects are defined as follows [51]: *Size effects are deviations from intensive or proportional extrapolated extensive values of a process which occur, when scaling the geometrical dimensions.* By this definition, size effects are defined as the deviation from the normal, e.g. expected behaviour, of a system. The scaling is further confined to geometrical scaling, which is probably the most important especially for micro technology. Intensive values are—according to the common definition in physical chemistry—system parameters like temperature or pressure and material parameters like density; they do not change with the volume of the sample. Extensive values are volume-dependent parameters like mass. The change in extensive value, e.g. the mass of a sphere, with changing size has to be considered when calculating the size effects. A proportional extrapolation simply means that one has to consider that an increase of a factor of 10 of the radius r of a sphere will increase the mass by a factor of 1,000, as the volume (and the mass, constant density provided) is proportional to r^3 . The definition also implies that a size effect can be quantified as a ratio (even a difference could be used) between the value x_1 and x_2 .

A systematic approach towards size effects was developed and introduced [50]. This system is based on three main categories of size effect, which in turn are subdivided in subgroups. Having such a system can help analyze existing effects in order to determine which mechanisms are behind a particular behavior.

In Fig. 2.1, the three main categories of size effects are shown. The name of the groups are given by the feature that is held constant during scaling and which is also responsible for the occurrence of the size effect. This method of addressing the different groups of size effects already shows that the scaling process will not change all parameters, since some will be held constant.

At first glance it might appear surprising that a constant density should be the source of a size effect. The common basic principle of those effects is based on the fact that a constant density of features (e.g. voids) also means a constant distance

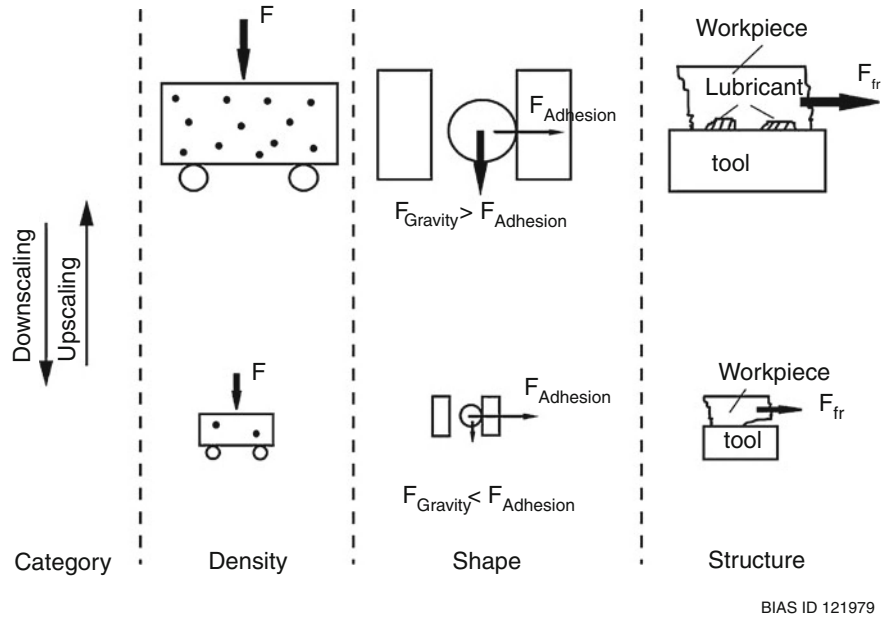


Fig. 2.1 Main categories of size effects, according to [51]

between them. If the sample size is large compared to that distance, the probability of the existence of such a feature in a volume element of the sample will be high. On the other hand, if the sample size is of an order of magnitude equal to the mean feature distance, there may be samples showing no such characteristic feature, which in turn results in the different behavior of the part.

The second category of size effects are characterized by having a constant shape, which addresses the effects produced by the ratio between surface and volume. Many effects used in nano technology are based on the shape size effects, as the relative surface area increases strongly with decreasing size. Consider a catalyst which enables chemical reactions through intermediate reaction phases at its surface. A sphere of 1 mm in diameter has a total surface area of 3.14 mm². As the surface to volume ratio scales with a factor of 1/r, the total surface of the spheres with the same total volume but a size of 10 nm will increase by a factor of 10⁵ to 314,000 mm². The common principle of shape effects is that there are volume related values (in our example: the number of parts) and surface related values (in the example the total surface area) which determine the observed value. The relation between these different components changes with size, which leads to a dominance of either one or the other effect.

The size effects summarized in the third group show a certain similarity to the first category, the density effects. The common source of the structure effects is also the relation between the sample size and the length, this length not being a mean distance between evenly distributed features but a discretely localized

distance defined by the atomistic or engineering structure of the workpiece. Again, the size of this characteristic length is held constant, which gives rise to size effects when the sample size approaches the size of the characteristic length.

Returning to the subject of essential mechanisms for the occurrence of size effects, one can conclude that the main source of size effects is the fact that it is not possible to keep the relation between all relevant parameters constant when reducing the size of the parts. As an example, imagine the scaling down of a sheet metal forming process. If one reduces the sheet thickness the relation of grain size to sheet thickness will increase. To scale down every factor, the grain size could be reduced. While this would keep the relation between grain size and sheet thickness constant, the strength of the material would increase according to the Hall–Petch-relationship. The higher strength will have an impact on the elastic deformation of the tools among other effects. Therefore, the correct scaling of all relevant parameters is not possible, and size effects are inevitable. This also leads to a definition of a general mechanism of size effects; this is based on the fact that different subprocesses or force components will have a different relative importance at different sizes. The nature of the processes is that all subprocesses and forces exist in all realizations of different sizes, but depending on the size certain aspects are dominant.

In order to show size effects, experiments with samples of different sizes are necessary. This results in the need for a nomenclature or definition for the size. There is a straightforward designation for size derived from the theory of similarity. To address a specific size of a part, a value called a ‘scaling factor’ (often λ is used as a variable, here Λ is used) with no dimension is taken. If we take Λ as the value, $\Lambda = 1$ would designate the ‘normal’ size of the part, e.g. a sheet thickness of 1 mm, while $\Lambda = 0.1$ would refer to a sheet of 100 μm in thickness, as it is intended that the ‘actual value = Λ times normal value’. This method to designate the absolute size of a realization has the advantage that Λ can be used as a variable (independent from features like length, surface area, volume or time) in formulas and graphs, if trends with variation of the size are shown. On the other hand, some information is lost using such relative values. One always needs calibration with respect to the ‘normal value’. In the given example the information ‘ $\Lambda = 1$ means sheet thickness = 1 mm’ describes this normal value. Without this information about the normal value it is difficult or impossible to know in which case size effects occur. Due to this, throughout this book the nomenclature introduced in [53] of ‘size<nn>’ will be used, where ‘<nn>’ is the initial sheet thickness or the rod diameter in μm . As an example, ‘size20’ designates a sheet with a thickness of 20 μm or a wire with a diameter of 20 μm . The example given above (a realization of $\Lambda = 0.1$ with a standard value of 1 mm) would be identified by ‘size100’. Using ‘size<nn>’ for designating a workpiece realization always contains the information that samples are produced by correct scaling, e.g. reducing as far as possible all relevant features by the same factor.

As the material behavior under plastic deformation is essential for micro forming (and also for the explanation of size effects), the grain size, and especially its dimension relative to the sample thickness, plays an important role. Due to that,

a variable κ , which is called ‘grain number’ is introduced. The grain number κ is defined as the (average) number of grains (having a grain size of l_G) along the thickness s_0 of the sample, e.g. the initial sheet thickness:

$$\kappa = s_0/l_G \quad (2.10)$$

So, for a grain size of $10\text{ }\mu\text{m}$ and a sheet thickness of $50\text{ }\mu\text{m}$ a grain number of 5 is obtained. The grain number can become also less than 1, if the grain size, measured in the sheet plane, is bigger than the sheet thickness. A sheet microstructure with a grain number less than 1 is also called a ‘Tiffany structure’.

Very often it is argued that size effects are detrimental and must be avoided. This is not true, as will be shown by the examples in [Chap. 6](#) where processes are discussed which are enabled by size effects. The use of size effects can also be seen from the strength measurement of materials. This also demonstrates the large variety of size effects. Figure 2.2 is a summary of different work on size effects affecting the strength of materials. On the x-axis the absolute size of the parts or the relevant area covered by the measurement principle is given. Essentially for larger samples this is the rod diameter or sheet thickness of the samples. In small dimensions, this size is given by the indentation depth, as strength measurements on samples in the nanometer dimension are often made by indentation hardness measurements. It is worth noting that size effects appear in a wide range of sample sizes, covering 6 orders of magnitude. On the y-axis a more complex value is used. It is called effect strength and defined (see [50]) to demonstrate the significance of the size effect, e.g. how strong the effect changes the observed value. Essentially, the effect strength is the ratio between the change in the value (here the strength of the material) with change in the size, while per definition the effect strength is always a positive value. In some cases, ‘smaller is stronger’ is true, but there are also cases in which ‘smaller is weaker’ applies. In both cases the effect strength is positive. It is obvious that size effects concerning the strength of materials which occur in very small parts are much stronger than those occurring in large parts. This emphasizes the importance of the consideration of such effects in mmf.

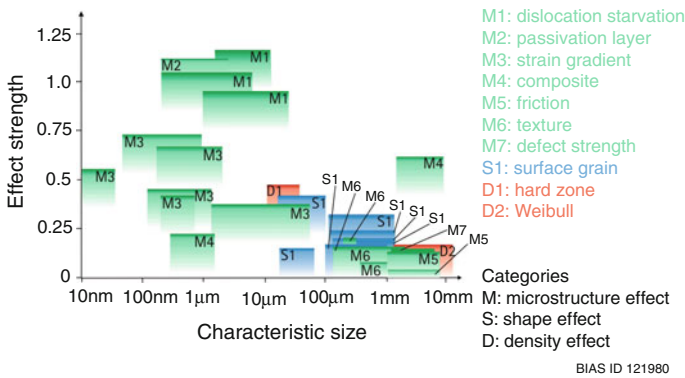
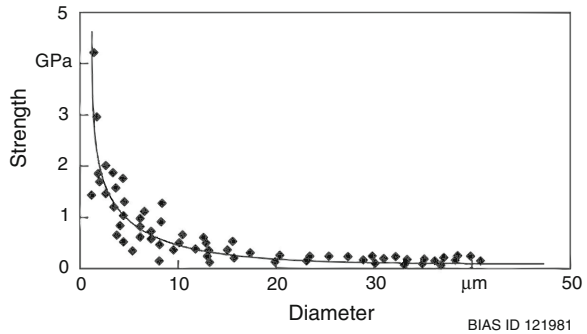


Fig. 2.2 Effect strength of different size effects affecting the flow stress, according to [51]

Fig. 2.3 Strength of copper whiskers, according to [51], data from [34]

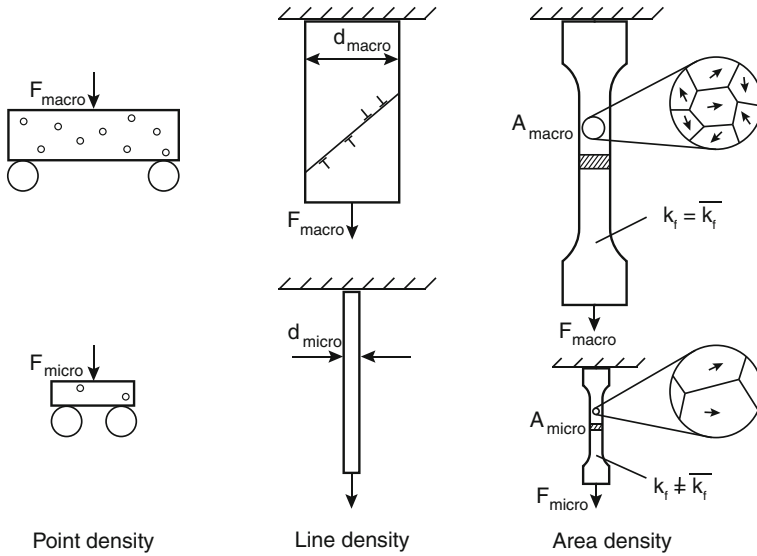


In order to demonstrate what size effects mean to the strength, some values from experiments using whiskers are replotted in Fig. 2.3. It demonstrates the strong increase in the strength of pure copper, normally known as a very soft material, up to values far above the strength of modern high strength steels. The effect is due to the fact that the sample diameter falls below the average distance between dislocations. Due to that the strength is not determined by the dislocation movement, but by the theoretical strength defined by the cohesion of the atoms within the lattice.

2.1.2 Density Effects

A density effect is a size effect which occurs during scaling when keeping the density of a feature constant. In particular, density effects will occur when the actual part size approaches the mean distance between the features. The density effects are subdivided by the three categories of the possible appearance of (microstructural) features. Those can appear as points (e.g. pores), lines (e.g. dislocation lines) or areas (e.g. grain boundaries). In the case of known size effects, there is a homogeneous distribution of such features though the volume of the part, while it cannot be excluded that a distribution of such features at the surface could also lead to density size effects.

The first possible feature shape shown in Fig. 2.4 is the point feature resulting in the point density effect. Those point elements can be pores in brittle ceramics or metallic welds, but also foreign atoms in a matrix lattice. The best known example of a point density effect is the so-called Weibull effect, which describes the size dependence of samples with low but not vanishing porosity. In large samples, pores near the surface give rise to stress concentrations and therefore they determine the strength of the part. Smaller samples may show no pores in the highly loaded region, because the density and therefore the distance between the pores is kept constant. Due to that, the probability of finding no pores within the critical region is increased, this results in the higher strength of smaller samples made from identical material.



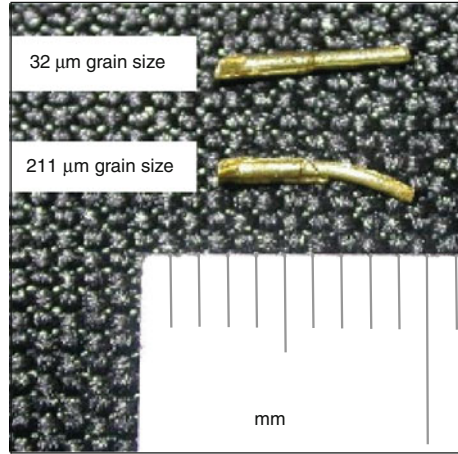
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Fig. 2.4 Subcategories of density effects

The line density effect can be observed in the case of copper whiskers, see Fig. 2.3. The line density is given by the initial dislocation density of the samples, which is typically 10^6 – $10^8/\text{cm}^2$, corresponding to a mean distance of 1–10 μm of the dislocations. Crystals with a diameter below 10 μm tend to have no (mobile) dislocations, resulting in a transition from deformation via dislocation movements to other mechanisms, which are dominated by the interatomic forces. This results in a very strong increase in the strength.

The third subgroup of the density effects in Fig. 2.4 is that of area density effects. The density of interfaces can be kept constant by maintaining a constant grain size. Sometimes the area density effects are determined by another layout of the experiments. Instead of decreasing the size of the samples at a constant grain size, the grain size is changed while the size of the samples is held constant. One should bear in mind that this is not the same. Despite the fact that the grain number can be changed by a layout such as the one in the experiments, the grain boundary density is not held constant, giving rise to additional effects like changes in strength according to the Hall–Petch-relationship. This must be accounted for if any forces or elongation limits due to plastic deformation are relevant. In some cases, experiments from such a layout can give interesting information. An example can be seen in Fig. 2.5 for extruded parts, with an initial grain number of 24 and 4. If only a few crystals exist along the cross-sections of the samples ($\kappa = 4$), the individual orientation has a strong impact on the local flow stress and in turn on the flow velocity. Those differences in the material flow are the source for the heavy distortion of the micro samples, which does not occur at a larger grain number ($\kappa = 24$).

Fig. 2.5 Workpiece distortion as a result of density effects [30]



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2.1.3 Shape Effects

Shape effects are such effects that occur if a process, assembly or part is carried out in a bigger or smaller realization with the same shape but different size. The easiest example is two spheres of different diameter. All these effects are based on the change in the surface to volume-ratio, which occurs during this kind of scaling. According to the general layout of the categorization of size effects, the reason for the size effects must be linked to the particular feature. Of course, in the experiments for density or structure effects the ratio of surface to volume will also change, but it is not essential for the observed size effect in these cases. On the other hand it cannot be excluded that density effects (and/or structure effects) and shape effects occur in parallel and sum up or eliminate each other. Experiments on size effects have to be checked carefully for such biasing effects.

Shape effects are also a very good example for the general size mechanism effects which are given above. This is due to the fact that shape size effects usually occur when the relation between surface and volume determined effects change, as the relative amount of the surface increases with decreasing size. Consider a cube with edge length l , then the volume V will be

$$V = l^3 \quad (2.11)$$

while the total surface area A is

$$A = 6 l^2 \quad (2.12)$$

The surface to volume ratio is

$$A/V = 6/l \quad (2.13)$$

e.g. it will increase $1/l$ with decreasing size.

Effects, forces or features which are dominated by the volume are the:

1. Mass of the part and all features which depend on it, such as
 - (a) heat content
 - (b) weight
 - (c) acceleration forces
 - (d) distances inside the volume, e.g. relevant for heat transfer by conduction
2. Amount of features in the particular volume, like
 - (a) number of grains
 - (b) number of pores
 - (c) amount of absorbed material

On the other hand, the surface is relevant for

1. Surface tension
2. Heat transfer (e.g. heat loss) by convection and radiation
3. Friction effects
4. Number of features at the surface, like
 - (a) grains with one boundary at the free surface
 - (b) amount of a adsorbed material
 - (c) electrical surface charge by electrons
 - (d) number of atoms, e.g. relevant for van der Waals forces.

Therefore, typically extensive variables are sources of shape size effects, while intensive variables might be the affected value. As an example, one could consider a hot forming process using raw parts heated to a homogeneous temperature in an oven. During the transport the parts will cool down by heat loss due to convection and radiation. The final temperature (an intensive variable) of the part just before the forming process will be determined by the heat content (one volume determined extensive variable) and the heat loss (surface determined extensive variables). With decreasing size the heat loss will become more and more significant, making the control of hot forming for micro parts very difficult.

Shape size effects, caused by a dominant volume or surface effect, might be the result either of concurrent or competing sub-processes. If the sub-processes principally point towards the same direction, e.g. they are concurrent, shape sum effects might occur. One of the most important shape sum effects for micro forming is the change in flow stress with changing size due to the influence of so-called surface grains [36]. As shown in Fig. 2.6, all grains of the microstructure will cause a positive contribution to the macroscopic measured flow stress. If we consider homogeneous grains with respect to chemistry and structure, the contribution will depend on the orientation of the individual grain relative to the loading direction and on the restrictions against deformation of the individual grain. This restriction is high at boundaries with other grains, while it is very low at free

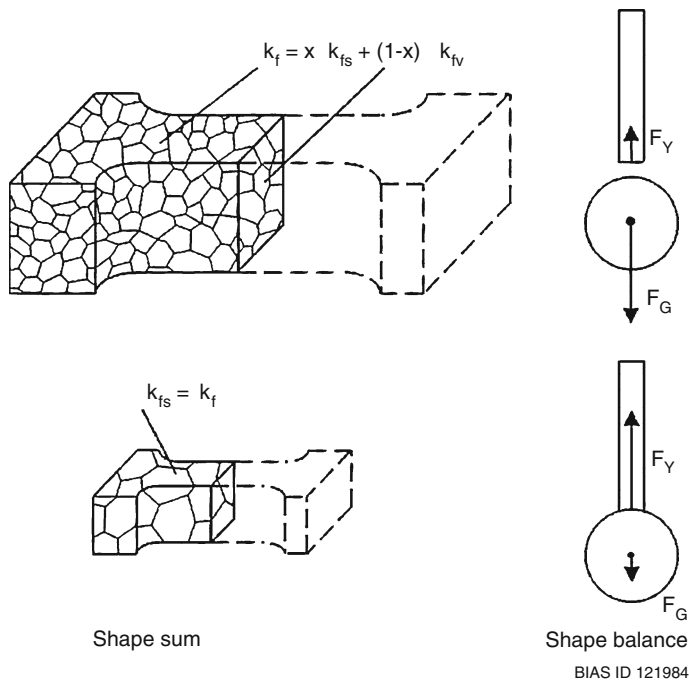


Fig. 2.6 Subcategories of shape effects

surfaces. The effects of the orientation may average out (if not the increase in scatter is a density effect, as the number of grains, not the surface to volume ratio, is relevant to that), while the influence of the softer surface grains will lead to a softer behavior of smaller samples, as the relative amount of the surface grains is proportional to the surface to volume ratio.

Shape size effects, where at least two forces or features act against each other, are called shape balance effects. The most prominent example of a shape balance effect is the problem of workpiece adhesion at grippers, which means some new strategies are necessary for part handling.

Figure 2.7 addresses that problem using a sphere as an example. There are three force components which hold the part at a gripper, while the gravitational forces act as the detaching force. The gravitational force is dependent on the mass of the part and is therefore a volume dependent force. The adhesion force, the van der Waals force and the electrostatic forces are dependent on the surface area and help keep the part on the gripper. While for large parts (far greater than 1 mm in the example shown in Fig. 2.7) the gravitational force is dominant, a decreasing size will lead to adhesion, increasing the necessary relative forces for detachment with decreasing size. If the size drops below 10 μm , even the small electrostatic forces can inhibit the detachment induced by gravitational forces. The role of adhesion is

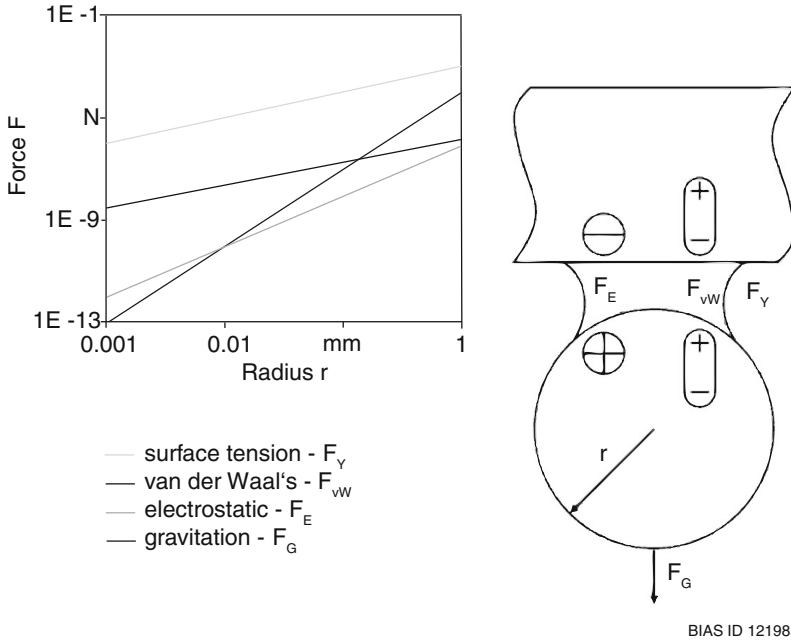


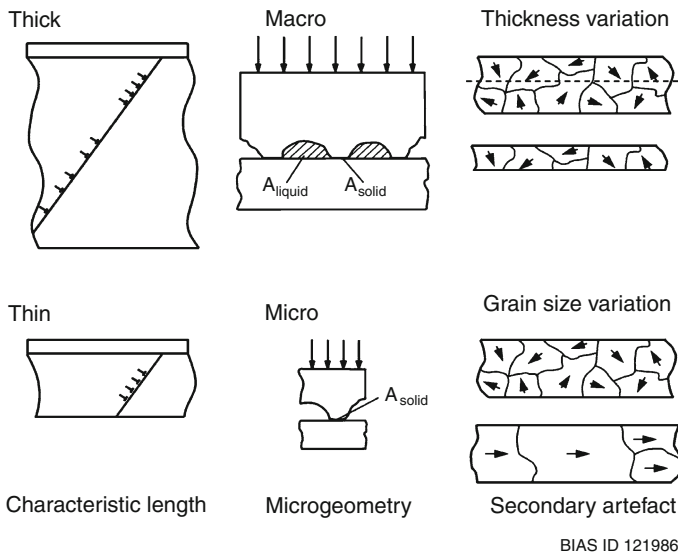
Fig. 2.7 Adhesion problem as a result of shape balance effects

very strong in the given example, as the existence of a liquid film between the gripper and part is assumed. In the case without a liquid layer, the adhesion forces will be much smaller.

2.1.4 Structure Effects

Within this context the term 'structure' summarizes the geometrical structure, e.g. the surface roughness or the geometry of a cutting edge on a cutting tool and the (atomistic) micro structure of the materials, e.g. the dislocation pile-up at a boundary or the texture of the grain orientation. Structure effects are those size effects which typically occur if the structure is held constant and a characteristic value of the part (e.g. sample thickness) approaches the characteristic length of the structure (e.g. length of a dislocation pile-up).

The first subcategory of structure effects is the characteristic length. A characteristic length is defined by the microstructure of the material, e.g. the length of the pile-up at a grain boundary, the width of a stress field around a precipitation (defined by a threshold value for the stress). The characteristic length in the strain gradient plasticity is also a characteristic value. For the example shown on the left hand side



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Fig. 2.8 Subcategories of structure effects

in Fig. 2.8, the length of the pile-up necessary to break through a hard surface layer is the relevant value. If the sheet thickness approaches this length of the pile-up, which happens at sheet thicknesses of some μm , the dislocation motion of any dislocation will be influenced by this layer, increasing the necessary stress for the movement of dislocations. This is measured macroscopically as an increase in flow stress. If the sheet thickness is increased, there will be areas in the sample which are outside the reach of the surface coating; e.g. the dislocation will move without any effect of the pile-ups at the surface layer, the sample appears to be softer.

It should be noted at this stage that, if one tends to argue that the Hall–Petch effect is also a structure size effect as it is also based on the consideration of dislocation pile-ups at grain boundaries, one should keep in mind that the Hall–Petch relation essentially describes the changes in strength with changing grain size for samples of constant size. Therefore, the Hall–Petch effect is not a size effect at all within the general definition of size effects, as the change in the macroscopic size of the workpiece is not necessary to obey the Hall–Petch-effect. Despite of this, experiments must be designed very carefully to avoid biasing effects from Hall–Petch effects in experiments on size effects. The latter is especially true for experiments where the grain size plays an important role and which are often analyzed using different grain numbers and constant thicknesses as explained at the end of Sect. 2.1.2.

Effects of the characteristic length cannot be avoided, as there is no possibility to change the characteristic length by the experimental set up. This is the case because the characteristic length is fixed by the nature of the material. In microgeometry effects this is not the case at first glance. Microgeometry effects are

also based on a constant feature size, but this feature is part of the (micro) geometry of the sample or the tool, e.g. the cutting edge radius of a cutting tool or the diameter and depth of a dimple on a sheet surface which is intentionally produced to improve lubrication during sheet metal forming. Principally, one could change the size of such features proportional to the size of the workpiece. Due to different reasons this is not done, e.g. as very sharp cutting edges are not stable and therefore not practicable for industrial use. In sheet forming the amount of lubricant is usually defined by amount per surface area, e.g. a smaller part automatically gets a smaller amount, in proportion to its surface. If the dimples, where the lubricant is stored, are also reduced, the excess amount of lubricant not stored in the dimples will increase. Due to this conflict it seems to be straightforward to keep the surface structure constant. Therefore, the same can happen as in characteristic length effects: If a characteristic size approaches the size of the microgeometric feature, size effects might be visible. One example for such a microgeometry effect is given in Fig. 2.8: Friction in lubricated (cold sheet) metal forming is determined by the efficiency of the lubricant. This can be increased by dimples, serving as reservoirs which carry the lubricant into the highly loaded contact zones. The characteristic size of the contact zone is size dependent, while the surface structure is held constant, which gives rise to a size effect. That size effect is used to explain the frictional behavior and its changes in micro forming.

The third subgroup of the structure effects is that of secondary artifacts. The definition or integration of the effects tries to include those that might be at the limit of ‘natural’ effects, but that occur in experiments which are linked with scaling effects. Effects, which cause a deviation from the expected behavior when scaling the size of the samples and which might be avoided by another experimental path, but are based on size dependent mechanisms, are taken as secondary artifacts. The term ‘secondary’ should express that it is not the direct link between two values in the experiment (like grain size and typical length of the part), which causes that effect, but that a longer chain of interconnected values is responsible for the effect. The best way to explain the nature of secondary artifacts may be an example. The result and the alternative of such a typical case are given in Fig. 2.9.

The intention of the experiments for the results shown in Fig. 2.9 was to answer the question: what will happen if the grain size not only approaches the sheet thickness, but exceeds it? According to the theory of surface grains, a shape sum effect, the strength of a sheet metal decreases when the sheet thickness is reduced at constant grain size, e.g. reducing the grain number. As shown in [47], the strength should become constant at the moment when the sample is of a single crystalline nature, e.g. below a grain number much lower than 1. Instead of the reduction in the sample thickness at constant grain size, some authors conducted experiments where they decreased the grain number by an increase in the grain size (at constant sheet thickness), derived by recrystallization and further annealing into larger grains. For $\kappa > 1$, e.g. there is more than one grain across the sheet thickness, the experiment showed qualitatively the same behavior as derived from experiments with a constant grain size and decreasing sheet thickness. That was one argument to accept it as a scaling experiment, despite the fact that the size

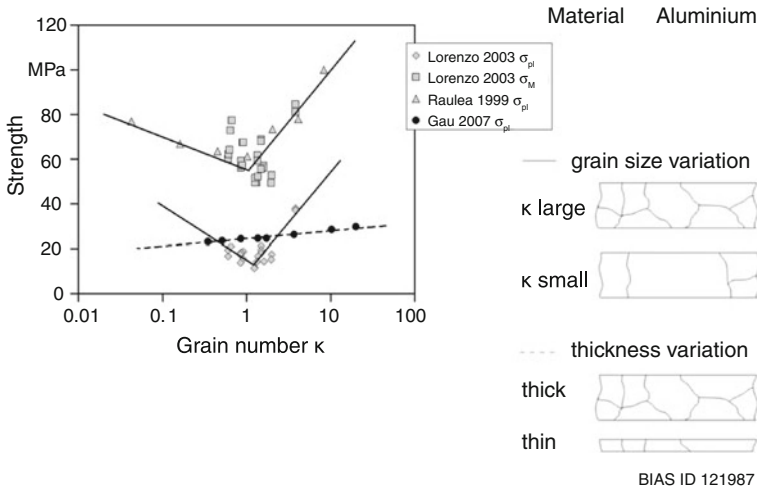


Fig. 2.9 Secondary artifacts in size effects on flow stress from various scaled experiments, data from [8, 12, 40], according to [51]

of the samples was kept unchanged. Surprisingly, samples with $\kappa < 1$ showed an increasing strength despite the fact that the Hall–Petch relation predicts a further decrease due to increasing grain size. The effect might be explained by a continuous change in the texture due to long recrystallization annealing. By that annealing a texture due to ternary recrystallization is developed, which has a ‘hard’ orientation of the grains towards the loading direction. As also shown in Fig. 2.9 this effect is avoided, if the sheet thickness is reduced at constant grain size. On the other hand, the effect might occur in a processing chain. Due to its indirect mechanism it is called a secondary artifact.

2.2 Tribology

2.2.1 Friction

Gerrit Behrens

2.2.1.1 Tribological System

In metal forming there is always an interaction between the workpiece and forming tool, caused by direct physical contact and relative motion. The result of this physical interaction is friction and wear. The science of interacting surfaces, which is related to friction, wear and lubrication, is called “tribology”.

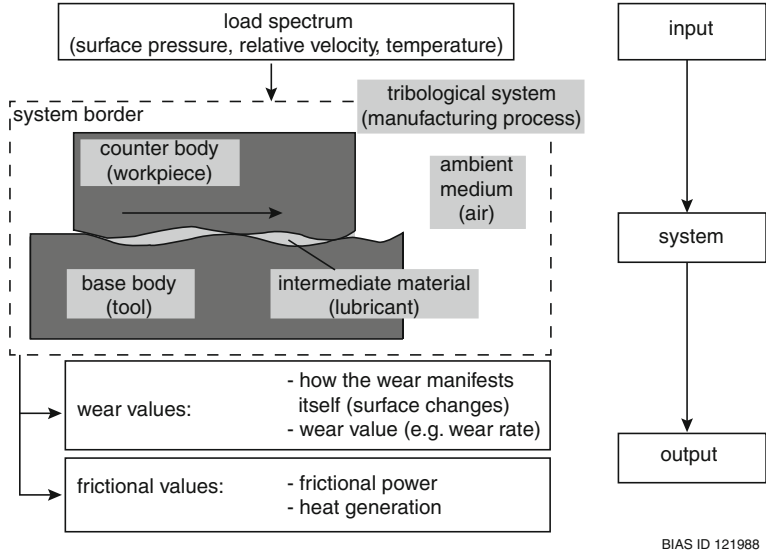


Fig. 2.10 Constitution of a tribological system [29]

The schematic diagram in Fig. 2.10 shows a two-body tribological system, which consists of the base body, counter body, intermediate material and ambient medium. The complete description includes the load spectrum as input values for the system, and the system response by means of wear and friction coefficients. In micro metal forming the base body is represented by the forming tool, the counter body is the workpiece and the intermediate material is the lubricant. Within a common manufacturing environment the ambient medium is air.

2.2.1.2 Characteristics of Friction

Friction is the force that resists relative motion between two bodies in contact. It is one of the most important forces in metal forming. The result of friction can, for example, be damage to the dies caused by wear, or the surface quality of the formed workpiece can be adversely affected. However, depending on the application, friction may also be beneficial. In metal forming in particular there are processes, such as rolling, where a certain amount of friction is essential for the correct operation of the process.

In sheet metal forming the resistance to tangential motion is usually described by the friction coefficient law, also known as the Amonton’s law of friction. In this model the coefficient of friction μ is the proportionality factor, which describes the ratio of the frictional force F_{fr} between two bodies and the normal force F_n pressing them together.

$$F_{fr} = \mu F_n \quad (2.14)$$

The locally effective frictional shear stress is defined as followed, where σ_n is the local normal stress:

$$\tau_{fr} = \mu \sigma_n \quad (2.15)$$

The tribological conditions in sheet metal forming are characterized by large contact areas and much lower surface pressures compared to bulk metal forming. If high normal stresses occur during forming, for example during bulk metal forming processes, the usage of the friction coefficient law is no longer valid. This is the case if the mean normal stress component is higher than the flow stress of the workpiece material. In that case the friction factor law is used instead. This approach connects the frictional shear stress with the yield stress in shear k using the friction factor m to describe the frictional behavior.

$$\tau_{fr} = mk \quad (2.16)$$

This model is best suited to characterize sticking friction or solid friction with an intermediate layer. The friction coefficient law is better suited for the boundary lubrication range, mixed lubrication, which is the most common case in metal forming, or hydrodynamic lubrication. Neither the friction coefficient law nor the friction factor law are therefore able to provide a correct description of friction in all cases. But the friction coefficient law alone is capable of describing the rheological behavior of a lubricant [31].

2.2.1.3 Testing Methods

There are many different testing methods for the determination of frictional characteristics, and the choice of a test method depends on its suitability for a given application. They can be divided into methods for sheet metal forming and bulk metal forming. While in sheet metal forming a biaxial stress state prevails, bulk metal forming is characterized by its triaxial stress state and much higher hydrostatic pressure. Since all testing methods and analytical friction laws are just simplifications of the real process, careful selection is necessary to prevent oversimplification or improper modeling. Sheet metal forming testing methods are mostly based on different kinds of stripe drawing tests or wedge drawing tests. In bulk metal forming the ring compression test is a common test procedure to determine the coefficient of friction μ or friction factor m . However, just a few of these tests have been used in micro scale experiments.

Most test procedures, such as the Duncan-Shabel test, require the use of special equipment and therefore a large number of very small devices. The resulting high costs prevent the application of many friction tests. This is why nearly all scaled experiments are based on the numerical identification of μ or m [51]. The ring compression test [2] or the double cup extrusion test [13] are the tests most

frequently used to acquire baseline data in micro bulk metal forming. Due to the necessity of numerical support these methods are indirect test methods. A possibility for the direct friction measurement in bulk metal forming is given by a compression sliding test [37]. For sheet metal forming, Hu determined the coefficient of friction for deep drawing from the maximum punch force. A further approach was based on an analytical model for a double deflection stripe drawing test [22]. FEM simulation was essential for all these investigations. Short descriptions of the mentioned test procedures are given below.

Ring Compression Test

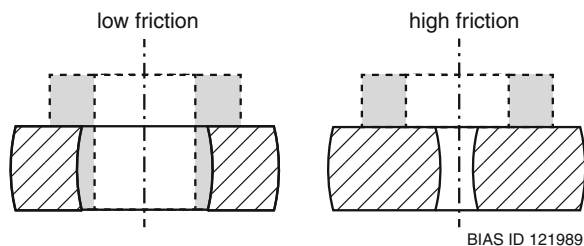
In this test, circular specimens with flat ends are compressed axially between two upsetting plates. Depending on the level of friction at the interface, the internal diameter of the ring may increase in the case of low friction or decrease if there is high friction. This change of inner diameter depends quite sensitively on friction, and the friction factor m can be easily determined from the change in the specimen's dimensions [2] (Fig. 2.11).

Double Cup Extrusion Test

The double cup extrusion test was applied as a friction test for the first time by [3]. For this test a cylindrical specimen is positioned in a die between two identical punches, one fixed and one movable. The downwards movement of the upper punch then forms the specimen material into two cups. The theoretical case of zero friction would yield two cups of identical height. With increasing friction the forming of the lower cup is prevented. Thus, small changes in friction can be detected by a significant change in cup heights, and the ratio of upper cup height to lower cup height is a sensitive measure for changes in friction [9]. However, if absolute values are required, these have to be identified numerically (Fig. 2.12).

The advantage of the double cup extrusion test compared to the ring compression test is the more realistic modeling of a typical extrusion process by applying larger strains and higher pressure between die and specimen. This best represents the actual situation in forward extrusion [13]. Because of the high

Fig. 2.11 Schematic ring compression test



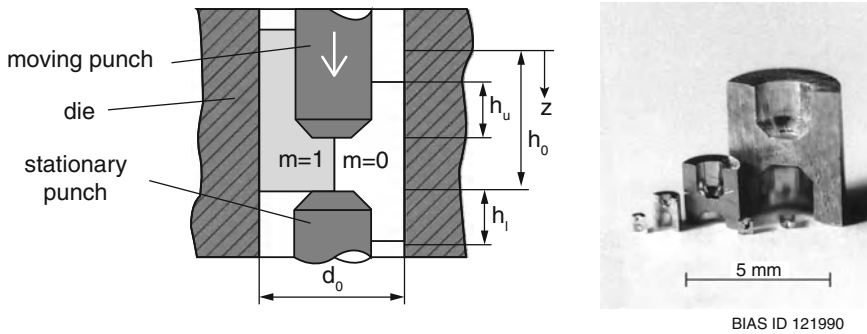


Fig. 2.12 a Setup of the double cup extrusion test and friction-dependent material flow. b Double cup extrusion specimen cut in half [9] (reprinted with permission from Elsevier)

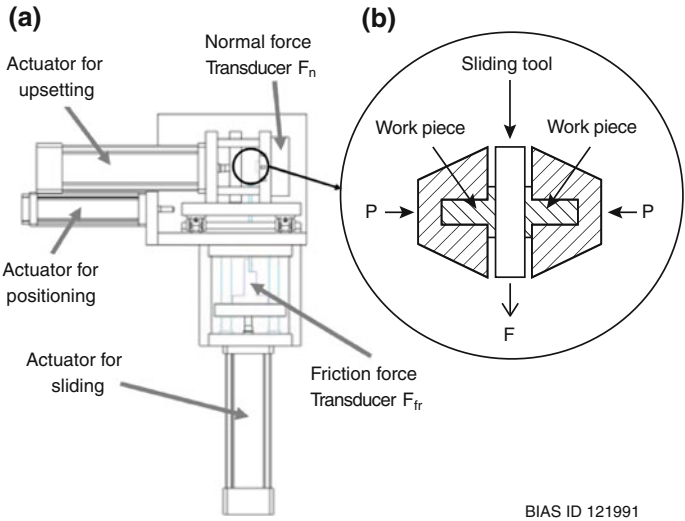
pressures and high normal stresses, the friction factor m is usually used as a characteristic parameter for the frictional conditions in this process.

Compression Sliding Test

In [37] a Micro-Tribo-Tester is presented allowing direct friction and lubrication tests in micro forming. This test is based on the symmetric upsetting of two workpieces against an intermediate sliding tool. First the specimens, which can be of different sizes and shapes, are upset to a preset reduction. This step is followed by the drawing of the sliding tool a predefined distance. Meanwhile the friction force is detected by a force transducer. The advantages compared to the double cup extrusion test [11] are the easy inspection of tool and workpiece surfaces caused by an open die geometry, and the continuous measurement of the friction coefficient over the complete sliding length and not just an averaged friction value for the whole sliding movement. However, this testing method is further removed from the realistic upsetting process conditions and is just a simulative not a process test (Fig. 2.13).

Double Deflection Stripe Drawing Test

Since friction is one of the most important influences on a deep drawing process it is essential to have available a reliable determination method for frictional behavior. The main advantage of the double deflection stripe-drawing test is its scalability, and it may therefore be applicable even for very small process dimensions [48]. In order to avoid excessive tangential pressure in the blank, which occurs in deep the drawing of circular blanks, the test procedure instead employs a rectangular blank. This allows the simplification of experimental and analytical procedures and additionally enables the calculation of friction

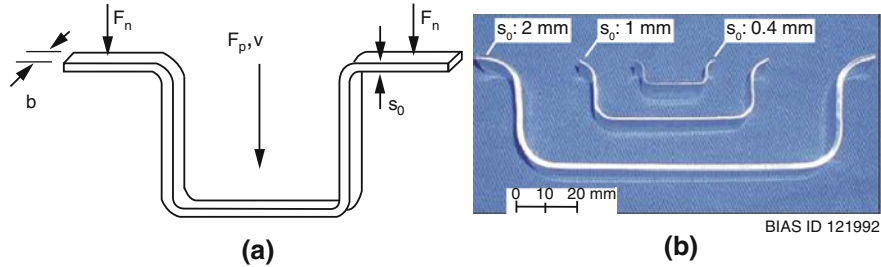


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Fig. 2.13 a Micro-Tribo-Tester. b Schematic compression sliding test (both [37])

coefficients. The principle of this friction test is shown in Fig. 2.14a, while drawn stripes in different process dimensions are shown in Fig. 2.14b.

While a stripe is drawn into the die the punch force is measured and the punch force versus punch travel can be obtained. The punch force consists of different bending forces and a friction force. Within a certain limited area of the process the bending forces can be theoretically calculated, so that the share of the friction force can be determined. Because the friction is dependent on the normal pressure, which varies in a great range during deep drawing, the friction force and the friction coefficient can be identified independently of the contact pressure by using a calculation model developed in [20]. The result is a contact pressure dependent friction function (see Fig. 2.15). Using this test method, and by comparing the friction functions from experiments in different dimensions, the influence of dimension on the tribology can be investigated. The determined friction functions



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Fig. 2.14 a Principle of double deflection stripe drawing test. b Drawn stripes in different process dimensions [48]

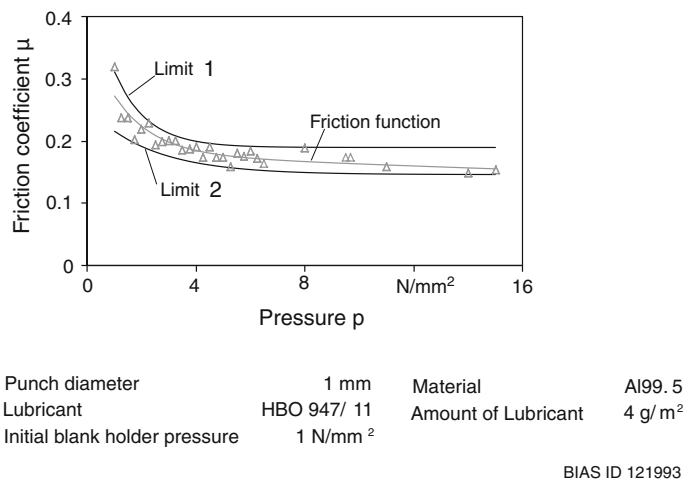


Fig. 2.15 Friction function based on friction coefficients calculated from experimental punch force versus travel curves [48]

can also be implemented in FEM-simulations to realize an improve model and taking the size dependency of the friction into account.

Due to the omission of tangential forces the double deflection stripe drawing test is just a simplification of the deep drawing process. A possibility to get information about friction in real deep drawing is presented in [49]. Here, the theory of Storoscchew is used to detect an effective average friction coefficient from the maximum punch force in micro deep drawing. This effective friction coefficient is not suitable to describe the frictional behavior for the whole process, but gives information about trends for changing friction when deep drawing process parameters are changed.

In Table 2.1 an exemplary overview over friction properties for different forming and testing methods is presented. It should be noted that the testing and forming conditions are completely different, which makes a direct comparison with each other very difficult.

2.2.1.4 Effects of Miniaturization on Friction

Many investigations in recent decades have shown a significant influence of scaling on friction in metal forming when lubricant was used. One of the first investigations of scaling effects in bulk metal forming, conducted by Pawelski, showed a reduction in the coefficient of friction with increasing dimensions. Independent of the testing procedure this trend could be shown for scaled ring compression test investigations by [10] as well as for studies carried out with the double cup extrusion test mentioned, conducted by [46]. Here the numerical

Table 2.1 Exemplary friction properties for different forming and testing methods

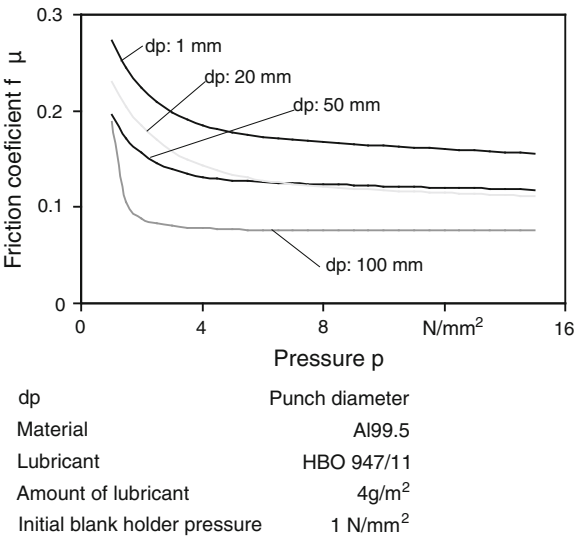
Forming type	Test method	Geometry	Material	Lubrication	Friction	Reference
Bulk	Double cup extrusion	$d = 4 \text{ mm}$	E-Cu57	Extrusion oil	$m = 0.01$	[46]
Bulk	Double cup extrusion	$d = 1\text{--}2 \text{ mm}$	E-Cu57	Extrusion oil	$m = 0.04\text{--}0.06$	[46]
Bulk	Double cup extrusion	$d = 4 \text{ mm}$	CuZn15	Extrusion oil	$m = 0.02$	[45]
Bulk	Double cup extrusion	$d = 2 \text{ mm}$	CuZn15	Extrusion oil	$m = 0.06$	[45]
Bulk	Double cup extrusion	$d = 4 \text{ mm}$	CuZn15	Extrusion oil	$m = \text{up to } 0.4$	[45]
Bulk	Cylindrical upsetting	$d = 4.8 \text{ mm}$	CuZn15	Dry	$m = 0.13$	[35]
Bulk	Cylindrical upsetting	$d = 1 \text{ mm}$	CuZn15	Dry	$m = 0.16$	[35]
Sheet	Strip drawing	$s_0 = 20 \text{ }\mu\text{m}$	Phosphor bronze	Dry	$\mu = 0.19$	[42]
Sheet	Strip drawing	$s_0 = 20 \text{ }\mu\text{m}$	Phosphor bronze	Dry	$\mu = 0.22$	[42]
Sheet	Strip drawing	$s_0 = 20 \text{ }\mu\text{m}$	Phosphor bronze	Dry	$\mu = 0.25$	[42]
Sheet	Deep drawing (max. F_P)	$d = 50 \text{ mm}$	Al99.5	Mineral oil	$\mu = 0.05$	[21]
Sheet	Deep drawing (max. F_P)	$d = 1 \text{ mm}$	Al99.5	Mineral oil	$\mu = 0.13$	[21]

identification of the double cup extrusion results yields a friction factor of $m = 0.01$ for specimen diameters of 4 mm and a drastic increase of $m = 0.04\text{--}0.06$ for diameters of 2–1 mm respectively, when the surface structure was held constant.

But not just in bulk metal forming is such a tendency of increasing friction with decreasing size observed. Results from double deflection stripe drawing tests by [48] show an increase in the coefficient of friction by a factor of 2.4 when scaling the process dimensions by 100 (see Fig. 2.16). In this investigation four different punch diameters of 100, 50, 20 and 1 mm were used. All geometrical parameters of the tools and necessary parameters of the specimens were scaled, according to the theory of similarity, by the same scaling factor. For instance the thickness and the length of the stripes had the same ratio to punch diameter in each experiment. Nevertheless, friction increased with decreasing punch diameter.

In further investigations the same trend observed for stripe drawing tests was reported for deep drawing [21]. Using the above-mentioned identification of the coefficient of friction from the maximum punch force, the average friction coefficient for scaled deep drawing processes could be achieved for punch diameters from 50 to 1 mm. While large diameters resulted in a coefficient of friction of

Fig. 2.16 Changing friction coefficients determined by a double deflection stripe drawing test [48]



BIAS ID 121994

$\mu = 0.05$ the deep drawing process with smallest punch diameter exhibited a significant increase up to 0.13. Independent of forming technology, a tribological size effect, showing an increase in friction with decreasing scale dimension, is observable whenever lubricated friction is used. The model of open and closed lubricant pockets is the only fundamental model available for the explanation of this size effect.

In the case of dry friction, such a significant tendency has not yet been shown. There were investigations into bulk metal forming as well as sheet metal forming, but while some find no size effects [35] others present increasing [15] or decreasing trends [42] for frictional behavior. The reasons for this behavior appear to be manifold and dependent on the actual individual case. There might be different explanations for the individual results, but it seems to be very doubtful that a general model explaining the differences in the coefficient of friction can be found. Therefore, a general size effect, as described for lubricated friction, does not exist for dry friction [51].

2.2.1.5 Lubricant Pocket Model

The lubricant pocket model is the only one well-established model to describe the general trend of increasing friction with decreasing dimensions in lubricated friction. It is based on the principle that lubricant is entrapped in asperities between the forming tool and workpiece. In so-called closed lubricant pockets the lubricant cannot escape as the pressure caused by the forming load increases. As a consequence a hydrostatic pressure develops and the lubricant bears part of the external

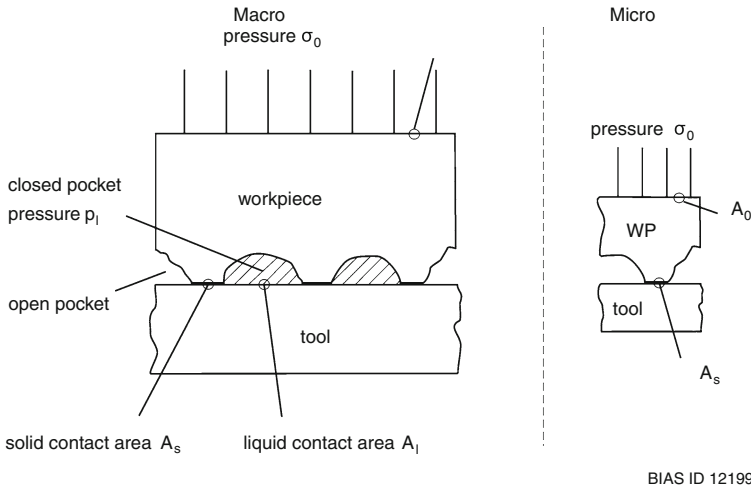


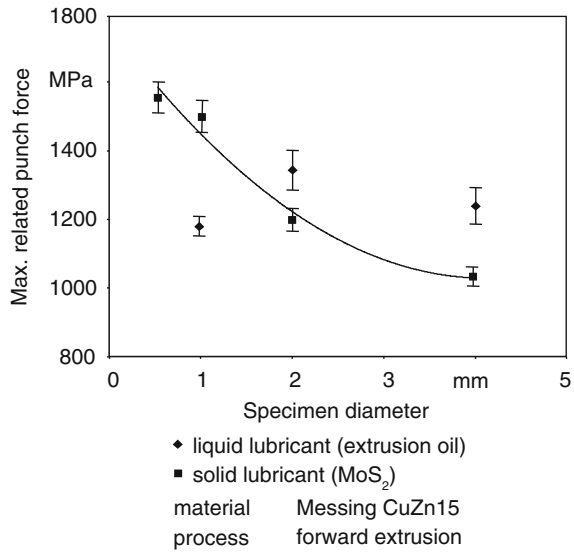
Fig. 2.17 Lubricant pocket theory for the coefficient of friction [53]

load which results in lower friction. This occurs when the pockets are fully sealed by the surrounding uninterrupted direct contact of the die and workpiece surface. Roughness valleys that are connected to the edge of the surface are not closed and cannot retain the lubricant. Therefore, the lubricant can pour out of the pockets and will not bear any load. These open lubricant pockets are not able to support the transmission of a load. The result is an increase in the normal pressure on the asperities and a higher surface flattening and therefore higher friction (see Fig. 2.17).

The area where open lubricant pockets exist is not restricted to the edge of the workpiece surface. Because the asperities are interconnected, the connections to the edge can reach far within the surface. Under the assumption of a constant surface topography the distance from the edge where open lubricant pockets can be found should remain the same, even when the process is scaled down. The consequence is an increasing share of open lubricant pockets when the process scale is decreasing [46]. If the dimensions are small enough no closed lubricant pockets will exist, which results in an increase in surface flattening and therefore higher friction. This assumption was confirmed by roughness measurements by [46]. A qualitative description of the lubricant pocket model is given by [13], and a quantitative explanation can be found in [9].

If it is true that the tribological size effect can be explained by the loss of lubricant in the open pockets, the size effect should not be visible if solid lubricant, which would be unable to flow and escape from open lubricant pockets, is used instead. This is confirmed by [14], where two extrusion tests with identical setups, using oil and MoS_2 as the solid lubricant, were carried out. While the results with oil clearly showed increasing friction with decreasing size such a trend could not be seen when using the solid lubricant (see Fig. 2.18).

Fig. 2.18 Size dependency of friction with different lubricants in forward extrusion (according to [14])



BIAS ID 121996

It can be summarized that the lubricant pocket model is valid for explaining the tribological size effect appearing in bulk and sheet metal forming whenever lubricant is used. The occurrence of this effect is due to the fact that the structure of the surface is held constant during scaling [53].

2.2.1.6 Velocity Dependence of Friction

Investigations into the tribological size effect in deep drawing with different punch velocities show the same tendency towards higher friction with decreasing size (see Fig. 2.19a), as predicted by the lubricant pocket model. However, this model implies that the coefficients of friction for macro and micro deep drawing should lie on the same Stribeck curve. This is not the case (see Fig. 2.19b) using a constant viscosity of the lubricant ($\eta = \text{const.}$). Nevertheless, it is known that the viscosity of lubricants is very strongly influenced by temperature.

There is good evidence that the temperature of the lubricant is higher during micro scale forming compared to macro scale forming. This is due to the fact that with increasing miniaturization less lubricant is kept in the micro tribological contact zone because more open lubricant pockets appear and the lubricant escapes from the asperities. Therefore, the temperature will increase when the same amount of heat per area is generated. Furthermore, frictional heat is generated as heat per area. Thicker sheet material, as used in macro deep drawing, acts more effectively as a heat sink than the thin blank material in micro deep drawing, and

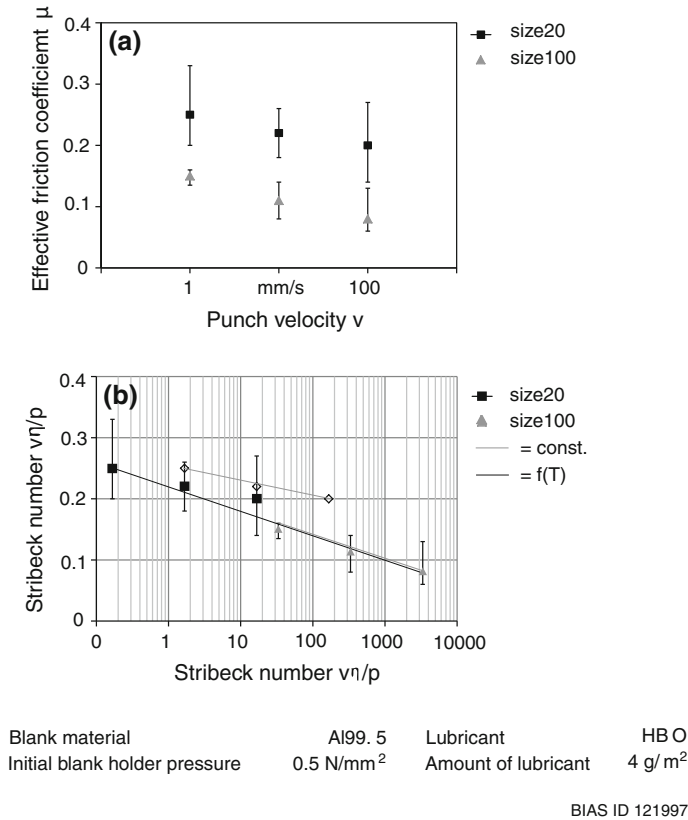


Fig. 2.19 **a** Change in coefficient of friction in deep drawing with different punch velocities and different scaling factors. **b** Structured size effect based on lubricant pocket model extended by a consideration of the temperature dependent viscosity of the lubricant (both [53])

therefore the heating of the lubricant is expected to be higher in a small scale process. The probably strongest effect might be the larger local contact pressure in micro range caused by the aforementioned reduced ability of the lubricant to support the transmission of the process forces. In conclusion, there is strong evidence that there is a significant difference in the temperature of the lubricant between a macro and micro forming process. However, a quantitative measure of the temperature increase is very difficult to obtain. So in Fig. 2.19b a qualitative estimate of 30 K is used, which is equivalent to a difference in viscosity of the factor of 10 ($\eta = f(T)$). Taking the temperature dependence of the viscosity of the lubricant into account, the values for the Stribeck number result in a singular Stribeck curve for both macro and micro deep drawing [53].

2.2.2 Tool Wear

Timo Kühnle

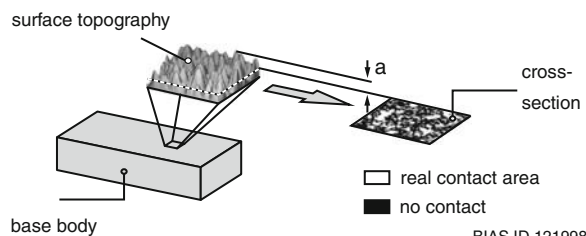
The Encyclopedia of Tribology gives a general definition for wear [26]: *Wear is an undesirable continuous loss of material from one or both of the surfaces of mating tribological elements due to relative motion of the surfaces.* Hence wear may be interpreted as the dissipation of mass and surface structure in contradiction to friction, which is related to the dissipation of energy.

The loss of material or surface structure can be attributed to four different wear mechanisms: adhesion, abrasion, surface fatigue and tribochemical reactions. These wear mechanisms are the result of the contact mechanics at the actual points of contact between the bodies at micro scale. A macroscopic way of looking at the surfaces is not suitable to describe these wear mechanisms, because even a macroscopically flat surface has many irregularities on microscopic examination. Each machining or forming operation leaves marks on the bodies' surface, e.g. grooves, scratches or furrows. Figure 2.20 illustrates the geometric properties of a surface topography with micro asperities.

The height of the asperities in the 3D-view is exaggerated to enable a clear visualization. From this point of view it is obvious that the imagination of a full-area contact between base- and counter-body does not reflect reality. Only a fraction of the bodies' total surfaces are in contact with each other. This portion is called the real contact area. Between the remaining surface portions there is no direct contact. For a given surface topography, the real contact area is a function of the contact pressure. The higher the contact pressure the greater the penetration depth a into the topography, the greater the real contact area. A cross-section through the topography reveals for a given penetration depth the real contact area [9]. The surface roughness is the reason for an inhomogeneous contact pressure distribution between the base and counter body at micro scale.

Via the in situ monitoring of the real contact area during the flattening of a single ideal asperity it is possible to quantify the relationship between the real contact area ratio and the contact pressure [54]. The sketch in Fig. 2.21a shows the ideal asperity with the geometry of a square-based pyramid before, during and after flattening. The corresponding graph in Fig. 2.21b shows the evolution of the ratio between the real contact area (RCA) and apparent contact area (ACA) with

Fig. 2.20 3D-analysis of the surface topography and estimation of the real contact area



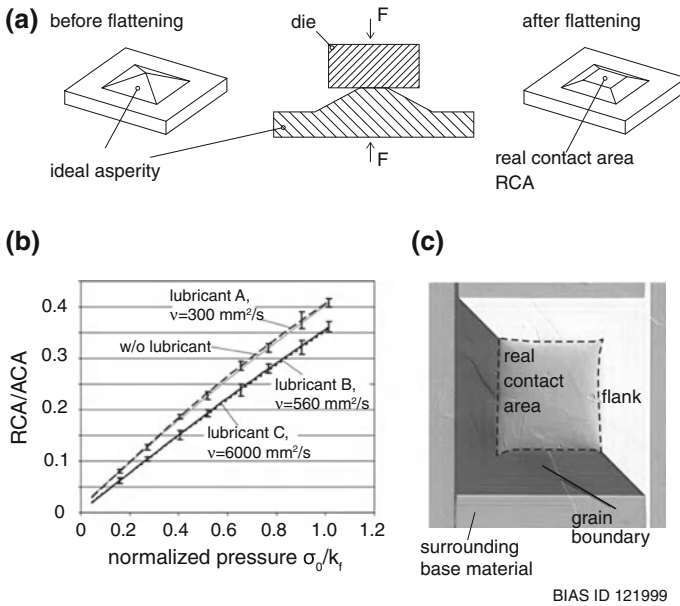


Fig. 2.21 **a** Flattening behavior of an ideal asperity. **b** Evolution of the ratio between RCA and ACA with respect to the normalized pressure. **c** Top view of the ideal asperity after flattening [54] with kind permission from Springer Science and Business Media

respect to the normalized pressure. In this example, the ACA is equal to the base area of the pyramid, which is $120 \times 120 \mu\text{m}^2$. The truncated pyramid in Fig. 2.21c is the result of the forming process under the boundary lubrication condition (BLC), whereby the RCA is enclosed with the dashed line. The BLC of a single asperity indicates the fact that initially there are no closed lubrication pockets in the contact zone. Nevertheless, the type of liquid lubricant has a significant impact on the evolution of the RCA ratio, even though there are no closed lubrication pockets. The reason for this behavior is connected with the formation of 2nd order closed lubrication pockets at high loads due to the roughening of the pyramid flanks. At low loads, the supporting effect is connected with a change in the lubricant properties. In general a small RCA ratio favors low friction and therefore is desirable for most forming processes.

In Fig. 2.22, the four basic wear mechanisms at the contact between microscopically rough bodies are shown. Abrasion and surface fatigue are mechanical dominated wear mechanisms. Adhesion and tribochemical reactions are driven by chemical–mechanical interaction within the tribological system.

Abrasive wear is characterized by the indentation of a hard counter body into the soft base body. The penetration with simultaneous relative motion leads to so-called micro plowing, micro cutting and micro cracking. During ideal micro-plowing the material is not removed from the base body but displaced to the sides of the groove. Therefore, the resulting wear volume is zero. Micro cutting is

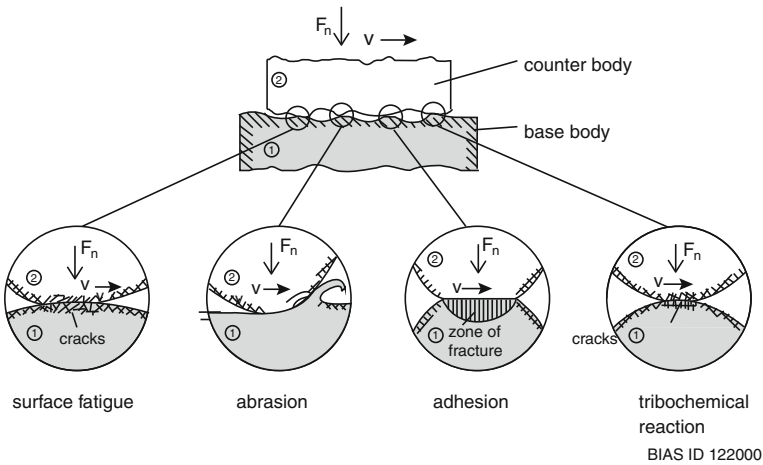


Fig. 2.22 Basic wear mechanisms from a microscopic point of view [5] with kind permission from Springer Science and Business Media

characterized by the generation of a small chip, which is removed from the base body. In this case, the wear volume is equal to the groove volume. Micro cracking means that mechanical load in the contact zone results in cracks in the surrounding area of the groove. The consequence is the formation of debris and the wear volume is greater than the theoretical groove volume. A good protection against abrasive wear is a high hardness of the base material, because a greater material hardness is associated with a decreasing depth of penetration of the counter body.

Surface fatigue leads to cracking beneath the surface of the base body during continuous operation and cyclic loading. The event of crack formation is followed by the phase of crack growth. The damage occurs suddenly when the critical crack length is reached after a certain number of cycles. The result is the delamination of the damaged boundary layer. Surface fatigue mainly occurs when there is a rolling contact between the base body and counter body. A good example of classic surface fatigue is the pitting of ball bearings or gear wheels.

Another wear-mode, which has a strong relationship with the delamination of boundary layers, is the so-called “eggshell-effect” or “thin ice-effect” of thin film coatings on tools. The effect occurs when a thin and hard but brittle layer on a yieldable base body is overloaded. In this case, the supporting effect of the substrate is too small. Once the yield stress of the brittle coating is exceeded, a crack perpendicular to the surface results from the bending stress in the layer. Different to classic surface fatigue, the crack initiation event is not related to cyclic loading. However, the similar crack growth mechanisms and debris formation mode due to delamination justifies the assignment of the eggshell effect in this category of wear mechanism.

Adhesive wear of metals is related to the formation of adhesive junctions between the base- and counter body. The forming mechanisms of these junctions

are comparable with the cold welding or sticking of material at micro scale. The separation of the connected asperities due to the relative motion of the surfaces indicates the actual adhesive wear event. In general, the material separation occurs not in the original contact area. Instead, a small proportion of the weaker material is transferred to the opposing body. According to [44] the formation of adhesive wear is subdivided into the following steps:

1. Deformation of surface asperities.
2. Removal of surface films (e.g. oxide layers).
3. Formation of the adhesive junction.
4. Failure of the junction and transfer of the material from the weaker body to the stronger body.
5. Modification of the transferred fragments.
6. Removal of the transferred fragments from the surface and formation of loose wear particles.

Direct contact between the base- and counter body is the precondition for adhesive wear. Therefore, the best prevention against adhesive wear is the separation of the sliding bodies with a lubricant. If direct contact is unavoidable, a weak tendency to build an adhesive junction reduces adhesive wear mechanisms. The tendency for the formation of adhesive junctions and their strength depends on the material combination within the tribological system. It is reasonable to distinguish between similar and dissimilar material combinations.

In the case of similar material combinations strong adhesion can occur. An example is the metal forming of iron base alloys, where both the tool and the workpiece are made of materials with similar chemical characteristics. The adhesive properties of similar material combinations are related to the lattice structure of the material. An experimental investigation in [43] reveals that metals with hexagonal closed packed lattices show less adhesion than metals with face-centered cubic lattices. It is believed that the reason for this behavior is connected with the different number of slip systems in the corresponding lattices. The first step during adhesive wear is the plastic deformation of surface asperities, which is attended by an increase in the RCA. Therefore, a metal with fewer slip systems is more brittle and the RCA is smaller in comparison to a ductile lattice with many slip systems. The result is a low tendency to adhesive wear.

A method to avoid adhesive wear is the application of dissimilar material combinations. The chemical bonding character of the materials is crucial for adhesive wear, especially for the formation and strength of adhesive junctions. To avoid adhesion with a metallic workpiece, the surface of the tool should not have the same bonding character as the workpiece material [19]. Anti-stick layers or coatings with ionic or covalent bonding characters reduce the tendency to adhesive wear with materials which have metallic bonds. An example for a coating with ionic bonding character is ZrO_2 . A practical example for the application of coatings with covalent bonding character is a diamond like carbon (DLC) layer on a deep drawing tool [24]. In comparison with an uncoated die with lubricant,

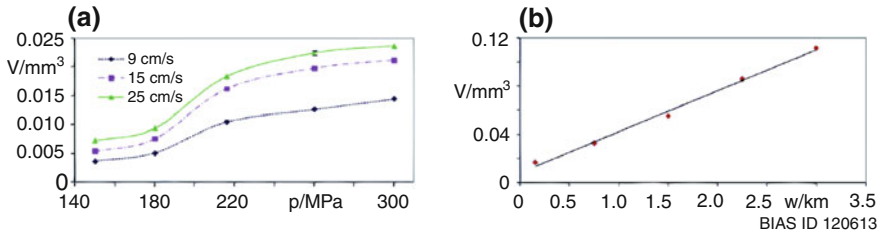


Fig. 2.23 **a** Influence of sliding speed and contact pressure on wear volume V at constant sliding distance $w = 300$ m. **b** Influence of sliding distance w on wear volume at constant velocity $v = 15$ cm/s and contact pressure $p = 260$ MPa [4] reprinted with permission from Elsevier

a DLC-coated die without lubricant reduces the punch force during forming. According to Hu, dry friction between a blank and coated tool is smaller than the friction of the uncoated tool with lubricant. Furthermore, no damage of the DLC-coatings due to eggshell effect, cracks or spalling was observed. These results show that a tool with DLC-coating have a great application potential for forming without lubricant.

Wear due to tribochemical reaction describes the phenomenon of chemical reactions between the base- or counter body with the intermediate material or the ambient medium. The reaction with oxygen of the ambient atmosphere is often observed and connected with term oxidational wear. The tribochemical wear mechanism is driven by the interaction of the micro asperities. These micro contact zones are subject to a severe temperature load due to the plastic deformation and the friction stresses in general. The high temperature increases the reactivity of the materials. Furthermore, the constantly repeated deformation removes already existing reaction inhibiting layers. In this way, constantly new reactive surface zones are generated. The effect of the tribochemical reaction on wear rate depends on the properties of the newly formed oxide layer. Initially the reaction products remain on the surface, but the thickness of the tribochemical reaction layer grows over the course of time. On the one hand, the brittle oxide products may be separated from the surface and form loose debris after reaching a critical layer thickness. In this case new debris will constantly be generated, which means a constant wear rate. On the other hand, the growth of the oxide layer will eventually stop if the reaction layer is permanently adhered to the surface. The result is a stable oxide interlayer and due to the permanent separation of the base- and counter body and further oxidation wear can be prevented.

There are various test setups to assess the impact of the input parameters of a tribological system on the wear rate or wear mechanism. The most important mechanical input values for a given tribological system are the contact pressure p and relative velocity v between the workpiece and tool. The result of a wear test is a wear rate diagram of so-called wear mechanism maps for the corresponding tribological system. For example, the wear rate diagram in Fig. 2.23a shows the

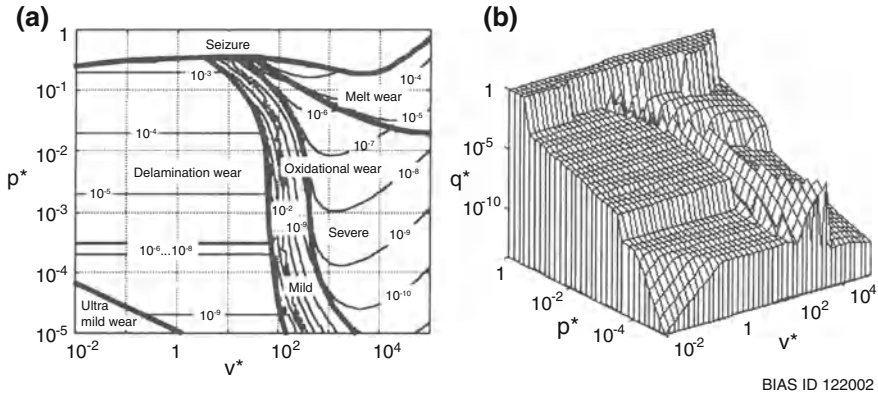


Fig. 2.24 **a** Lim and Ashby's wear mechanism map [32]. **b** 3-dimensional plot of the wear map [39] reprinted with permission from Elsevier

impact of sliding velocity and contact pressure on the wear volume V at a constant sliding distance of 300 m [4]. Within this narrow parameter range it is possible to conclude that the greater the contact pressure or sliding velocity, the higher the absolute wear volume. However, it should be noted that there is no linear relationship between the wear rate and contact pressure or sliding velocity. In contrast, the relationship between wear volume and sliding distance in Fig. 2.23b shows a linear behavior, which can be described by Shawn's wear law

$$V = q \cdot w \cdot F_n \quad (2.17)$$

where q is the wear rate, w is the sliding distance and F_n is the normal force [41].

With wear mechanism maps it is possible to show the transition of different wear mechanisms as a function of the input parameters of the tribological system. In contrast to the wear rate map in Fig. 2.23a, for wear mechanism maps it is necessary to vary the normalized input parameters over several orders of magnitude. Figure 2.24a shows a wear mechanism map for steel [32]. This map describes the unlubricated wear behavior of steel on steel in a pin on disk configuration. The map labels the dominant wear mechanism in the corresponding pressure–velocity regime. Furthermore, the predicted normalized wear rates q^* are superimposed over these fields. The transition from one wear mechanism to another is connected with a volatile variation of the wear rate. The 3-dimensional plot of the normalized wear rate in Fig. 2.24b illustrates the relation between wear mechanism and the corresponding wear rate [39].

2.3 Formability

Hendrik Flosky

2.3.1 Measures for Formability

Formability is a critical factor in metal forming processes. It depends on the chemical composition of the material, its crystalline structure, the temperature, and on the heat treatment. Also, size effects have an influence on formability (see Sect. 2.1). Formability can be measured with, e.g., a tensile or a compressive test (see Sect. 3.3.1). For formability, the strain is especially relevant, since the material needs to elongate uniformly without undergoing necking or fracture during the forming process. The strain is defined by

$$d\varepsilon = \frac{dl}{l_0} \Rightarrow \varepsilon = \int_{l_0}^{l_1} \frac{dl}{l_0} = \frac{l_1 - l_0}{l_0} \Rightarrow \varepsilon = \frac{\Delta l}{l_0}, \quad (2.18)$$

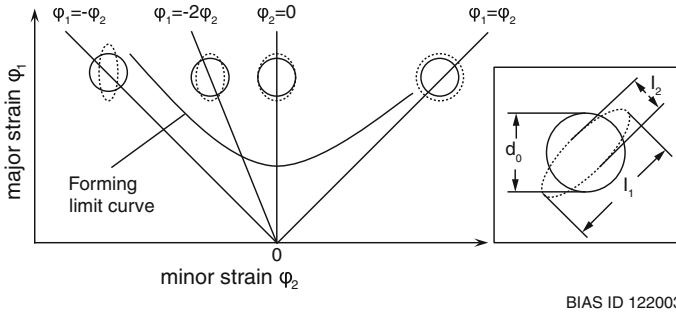
where l_0 denotes the specimen's original length, the length during the deformation is l_1 , and the length variation is dl . To calculate the technical strain ε , the right-hand side $\left(\frac{dl}{l_0}\right)$ of the equation has to be integrated over the deforming path. The plastic deformation can also be shown in a flow curve (see Sect. 3.3), where the strain is defined as the logarithmic degree of deformation φ . It can be explained, e.g., with the uniaxial tensile or compression test (see Sects. 3.3.1 and 4.2). The tensile test stripe has an initial length of l_0 , and l_1 is its length after elongation. During the deformation, the length variation is dl and the momentary length l . In order to obtain φ , the right hand side $\left(\frac{dl}{l}\right)$ of the equation has to be integrated over the deforming path:

$$d\varphi = \frac{dl}{l} \Rightarrow \varphi = \int_{l_0}^{l_1} \frac{dl}{l} = \ln l_1 - \ln l_0 \Rightarrow \varphi = \ln \frac{l_1}{l_0}. \quad (2.19)$$

In a parallel way, the formability for bulk metal forming can be measured with the upsetting process, where the initial height is h_0 and h_1 is the height after compression:

$$\varphi_h = \ln \frac{h_1}{h_0}. \quad (2.20)$$

However, the specimen has to remain cylindrical without getting convex contours during the upsetting process to calculate the formability in the right way. This can be done with the right lubricant and the right specimen geometry (see Sect. 4.2).



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Fig. 2.25 Forming limit curve in a forming limit diagram according to [17]

Thin sheets and foils tend to undergo necking and fracture during the forming process because of high local strains in the material. Therefore, it is necessary to measure the formability of the material, including the points of necking and fracturing. Most industrial processes require parts without any necking, hence the point of necking is the most meaningful criterion of failure. A relevant approach in the research field of local strains in sheet metal forming was made by Keeler [27] and Goodwin [16]. Hasek joined their researches into a forming limit diagram (FLD), as shown in Fig. 2.25.

Formability can be visualized by different mechanical stresses: deep drawing ($\varphi_1 = -\varphi_2$), uniaxial tension ($\varphi_1 = -2\varphi_2$), plane strain ($\varphi_2 = 0$) and stretch draw ($\varphi_1 = \varphi_2$), shown in Fig. 2.25 as the deformation of the circle. The forming limit curve divides the forming limit diagram into two parts. Above the curve, the material necks or fractures and below, the material elongates uniformly. Hasek measured the deformation using circles, defining the major strain φ_1 (x-axis) and minor strain φ_2 (y-axis) to visualize the deformation of the circle by

$$\varphi_1 = \ln \frac{l_1}{d_0} \text{ resp. } \varphi_2 = \ln \frac{l_2}{d_0} \quad (2.21)$$

These strains can be calculated using the starting diameter d_0 (before deformation) and the ellipse diameters l_1 , respectively, l_2 (after deformation). The deformation depends on the mechanical stress. For example, in deep drawing ($\varphi_1 = -\varphi_2$), the circle turns into an ellipse, and in stretch drawing ($\varphi_1 = \varphi_2$), it turns into a larger circle. Another aspect of the formability analysis on thin metal sheets is that the two-dimensional stress state can be used to describe the stress of the small circles and it can be only used to measure linear strain paths.

A line net system is normally used to measure the deformation of the FLD and its points. These lines form circles, which are used to measure the deformation and are printed on the materials surface with, e.g., a photochemical process or laser polishing. However, the mechanical properties of thin metal sheets and foils are influenced by these processes. To avoid any influence on the mechanical properties, one solution is to apply a stochastic pattern, by spraying on the foil surface [6]. The foil with a stochastic pattern, also called a speckle pattern, is measured

with a charge-coupled device (CCD) camera system. Two CCD cameras record and analyze the deformation process, allowing a 3-D measurement. In Sect. 5.2.1, the result of such a 3-D measurement is shown. Figure 2.26 shows the bulge test including the CCD camera system.

During a bulge test, which can also be used to measure the flow curve, a sheet or a foil is deformed until it fractures. The left part (1) of the schematic depicts the specimen before and the right part (2) during the deformation. The pressure needed in the bulge test can be generated either by a liquid (hydraulic) or a gas (pneumatic). In [7], both systems are compared and in [52], only the pneumatic bulge test is used. Hoffmann compared in [18] the flow curves of Aero-Bulgetest, which is similar to a pneumatic bulge test, with the flow curves of a tensile test. The material used was pure copper (99.5 %) with thicknesses from 10 to 100 μm . He indicates that the advantage of the Aero-Bulgetest over the tensile test is the possibility of measuring the biaxial stress state of the flow stress and its reduced processing errors in calculating the flow stress with a higher precision. The minor strain of the bulge test can be varied by using dies with different elliptical aspect ratios. Every elliptical aspect creates a point in the forming limit diagram. A disadvantage of the bulge test is that only positive major strains can be measured. Hence it covers only the right side of the forming limit diagram, the biaxial tension.

On the other hand, the so called Nakajima test covers the whole forming limit diagram. The Nakajima test uses a rigid punch for deformation instead of hydraulic or pneumatic pressure, and different sample geometries are tested successively. Every specimen generates a point in the forming limit curve of the forming limit diagram. This allows different specimen widths, as shown in Fig. 2.26 (right). With this variation, the minor strain of the forming limit diagram can be controlled. The Nakajima test is used in [6] to measure the forming limit of thin foils (Al 99.5 and SE-Cu58).

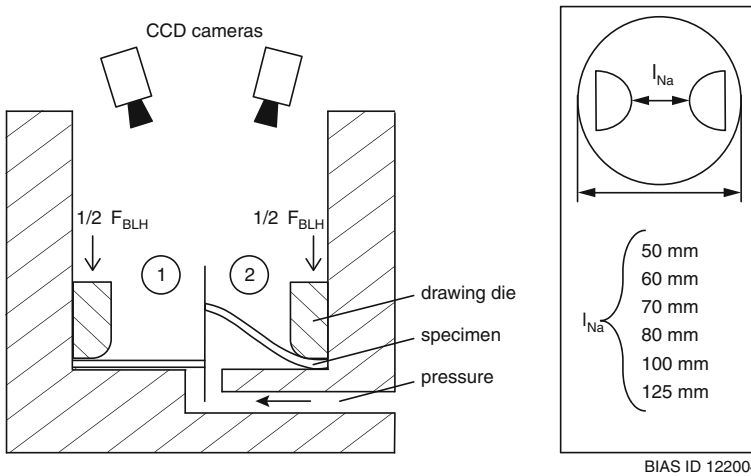
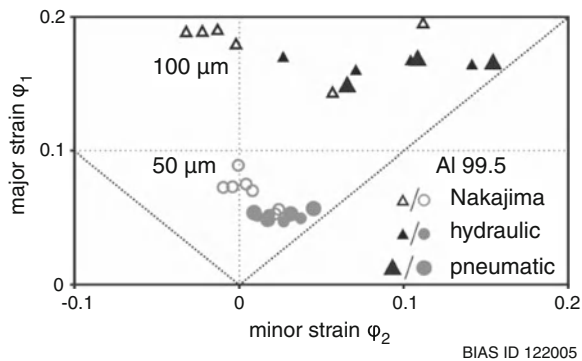


Fig. 2.26 The bulge test (*left*) and Nakajima sample geometry (*right*) according to [7]

Fig. 2.27 The hydraulic and pneumatic bulge test and the Nakajima test, sheet thickness 50 and 100 μm with permission from Springerlink [7]



By using different measurement methods, it becomes difficult to compare the results. Hasek [17] was the first to attempt a formability comparison with thin sheets. There are differences in the relation between the principal strains and the friction conditions between the methods. For example, the bulge test is frictionless and needs no lubricant. If the Nakajima test is carried out without a lubricant (or with an unsuitable lubricant), the results are incorrect and the comparison is more complicated [7]. In Fig. 2.27, three testing methods (pneumatic bulge, hydraulic bulge, and Nakajima) are compared using the forming limit diagram. The specimens material used is pure aluminum (Al 99.5) with thicknesses of 100 and 50 μm .

The results from the hydraulic and the pneumatic bulge test are similar. By comparing the results regarding the Nakajima test and both the hydraulic and the pneumatic bulge test (right side of the forming limit diagram), there is a higher variation between the values than the comparison between the pneumatic and the hydraulic bulge test.

2.3.2 Formability in Micro Metal Forming

In order to understand formability in micro metal forming, it is necessary to comprehend the structure of the materials, since this has a great influence on the micro structural size effect (see Sect. 2.1). The material's crystal structure is normally classified as either single crystalline or polycrystalline, as shown in Fig. 2.28 on the right side. A polycrystalline structure consists of many grains in every dimension, which leads to a uniform deformation and a high formability. A single crystal consists, as its name suggests, of only one crystal, and has a relatively low forming potential. An interesting and important aspect in micro metal forming is the behavior of a material which is in between polycrystalline and single crystalline.

Armstrong [1] was the first to describe the transformation from polycrystalline to single crystalline. A polycrystalline material can be described as a homogeneous continuum, and its material strength can be found by using the Hall-Petch relation

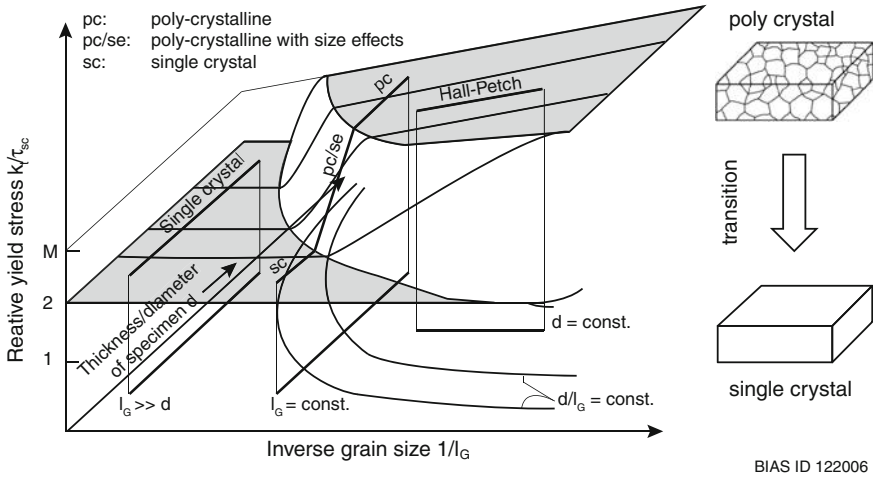


Fig. 2.28 Transition from poly- to single-crystalline structure [47]

(see Sect. 3.3.4). As shown in Fig. 2.28, when the inverse grain size $1/L_g$ and/or the sample thickness decreases, the material behavior changes from polycrystalline (pc) to single crystalline (sc). Between these states, the material behaves like a poly crystal with size effects (pc/sc), where the flow stress drops to a lower level. In addition, the change from polycrystalline to single crystalline includes an increase of the flow stress scattering.

There are different opinions in the literature about how many grains are in the cross section when the transition starts, e.g., [1] estimated the start to be at less than 20 grains. But Justinger demonstrated in [25] that the grains in the cross section of the deformed volume should be 50 or higher. The model also includes the scatter of the material strength. If the grain number is less than 50, the scatter increases.

Beside poly- and single crystalline, three other structures can be defined: Oligo crystalline (a), bamboo (b), and the Tiffany structure (c). The oligo structure has few grains, but more than one in the cross section and many grains within the plane. The bamboo structure is defined as having only one grain across the diameter, but many along the length. It has the shape of a bamboo rod, because the area where two grain boundaries are in contact deforms less than the area with only surface grain boundaries during the deformation.

The Tiffany structure, first mentioned in [53], has (like the oligo structure) many grains within the plane, but only one grain in the cross section, as shown in Fig. 2.29d. Figure 2.29d, e shows where the name Tiffany structure comes from. The surface of an etched specimen looks like Tiffany glass and has also only one grain in the cross section, as shown in Fig. 2.29f.

The different orientations of the grains are evidenced by the various grain colors, and these orientations have an influence on the mechanical properties. Single grains behave differently depending on their orientation and loading

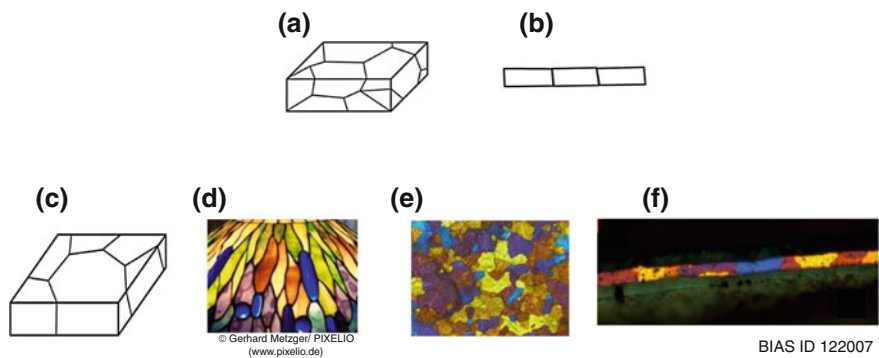


Fig. 2.29 Material structures. **a** Oligocrystal (bulk). **b** Bamboo (rod). **c** Tiffany structure (sheet). **d** Tiffany glass. **e** Etched structure (top view). **f** Etched structure (cross section 20 μm)

direction. As shown in Fig. 2.30, the strain of the foil (52.5 μm) during laser shock forming (see Sect. 5.2.2) has an inhomogeneous behavior. By comparing the real strain and the expected one, it is clearly seen that the strain differs not only in one direction. This confirms the Tiffany structure’s characteristic, where differently oriented grains have various loading directions and different strains.

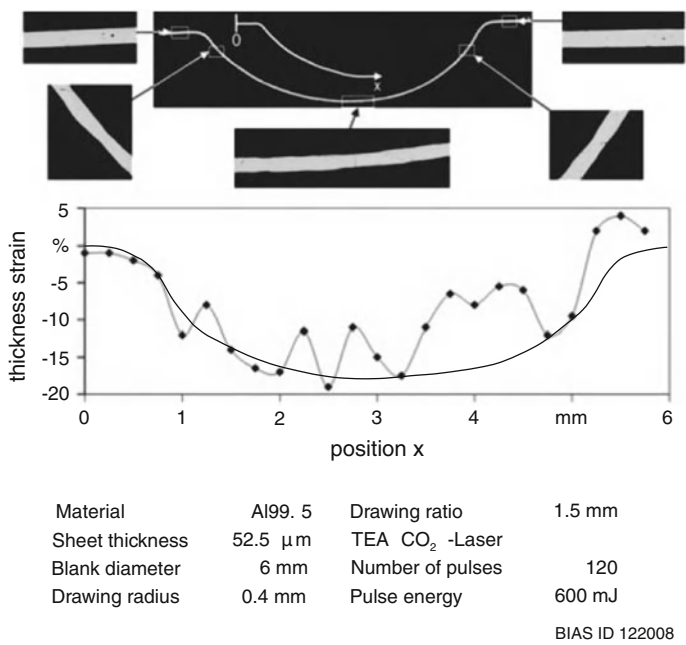
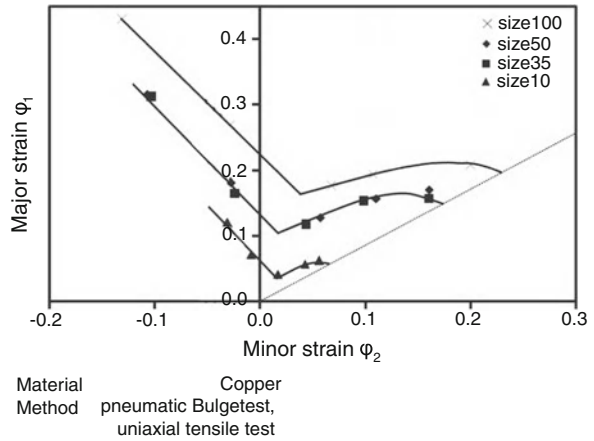


Fig. 2.30 Behavior of the strain in a laser shock forming specimen after [55]

Fig. 2.31 The pneumatic bulge test tensile test, sheet thickness 10, 35, 50 and 100 μm [28]



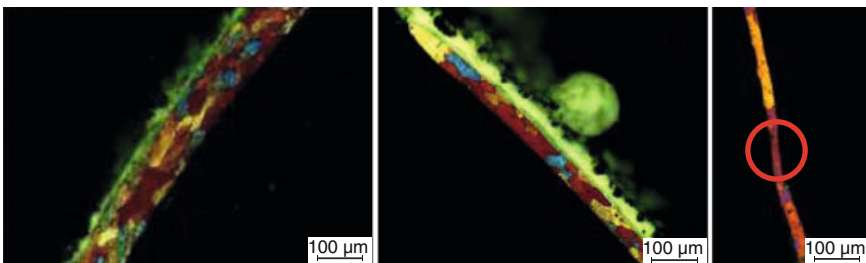
BIAS ID 122009

Various copper foils with thicknesses from 10 up to 100 μm are depicted in a forming limit diagram (Fig. 2.31). It can be seen that the thinner the foil, the lower the major strain and the lower the forming limit.

Another important aspect, which is not shown in Fig. 2.31, is that the forming limit curve cannot be generated as simply as in macro forming, since the variation of the values is much greater [23]. This is caused by the growing influence of the surface grains when the foil is thinner (see Sect. 2.1), until a Tiffany structure is reached. In [33], it was estimated that the elastic component varies more in thinner foils, which can also be explained by the growing influence of the surface grains (see Sect. 5.3).

The different grain orientations and the greater influence of the surface grains make formability prediction more difficult. But even in micro metal forming, a good prediction is essential because of the maximum sample size of only 1 mm in two dimensions, which means even a minor deviation could make the sample useless.

To visualize the influence of the Tiffany structure, Fig. 2.32 depicts its etched micrographs. Each picture shows a part of the fractured Al 99.5 specimens. As



BIAS ID 122010

Fig. 2.32 Micro structures of stretch drawn parts (Al99.5) with different thicknesses **a** (size100), **b** (size50) and **c** (size20) [23]

explained before, the different grain colors are due to the different orientations. The micrographs show how the foil thickness decreases. In Fig. 2.32c, the foil is thinner than the grain diameter. In the red circle, it can be clearly seen that the foil becomes thinner, because of the different grain orientation. Another critical aspect, which is similar to the sheet thickness, is the proportion between the grain size and the specimen thickness (see Sect. 2.1).

Hu carried out the same test using copper (E-Cu58) with size20 and size50, obtaining the same results as with aluminum. The variation of the values is large; the thinner the foil, the lower the forming limit. In [6], experiments were carried out with the bulge test and the Nakajima test: the formability decreases and the result variation increases when using thinner foils, as shown in Fig. 2.27.

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Chapter 3

Mechanical Properties

Hans-Werner Zoch

Latin

A	Area (mm^2)
A_G	Grain area (mm^2)
b_v	Absolute value of the Burger's vector (nm)
C_{th}	Thermal material resistance (MPa)
CDF	Cumulative distribution function
CDF_e	Cumulative distribution function of the endurance limit
c_{fa}	Concentration of foreign atoms (at.%)
d	Diameter (mm)
d_G	Grain diameter (mm)
E	Young's modulus (GPa)
f_{Debye}	Debye frequency (s^{-1})
F	Force (N)
G	Shear modulus (GPa)
GBR	Grain boundary resistance $\text{MPa}\sqrt{\text{mm}}$
GFE	Gibbs free enthalpy (J)
HM	Martens hardness (MPa)
K	Prefactor (MPa s)
K_B	Boltzmann constant (J/K)
K_f	Fatigue notch factor
K_t	Stress concentration factor
l	Length (mm)
L	Gliding path (nm)
$\bar{L}$	Mean distance between short-range obstacles (nm)
$m_{\text{th}} n_{\text{th}}$	Exponents
m_f	Strain rate sensitivity

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m_{of}	Orientation factor
m_{Sachs}	Sachs' factor
m_{Taylor}	Taylor's factor
m_{Weibull}	Shape parameter
N	Number of cycles
N_{f}	Number of cycles to failure
Pr	Probability
P_{rf}	Probability of failure
Q	Activation energy (kJ/(mol K))
R	Gas constant (kJ/mol)
R_{σ}	Stress ratio
R_{a}	Arithmetic roughness (μm)
R_{t}	Maximum height of profile parameter (μm)
R_{y}	Maximum peak to valley height (μm)
R_{z}	10-point height parameter (μm)
s	Sheet thickness/layer thickness/material thickness (μm)
t	Drop-down time (s)
T	Temperature (K or $^{\circ}\text{C}$)
T_{R}	Recrystallisation temperature (K)
T_{M}	Melting temperature (K)
t_{D}	Delay time of a glide dislocation in front of an obstacle (s)
t_{L}	Running time of a glide dislocation (s)

Greek

α	Heat transfer coefficient (W/($\text{m}^2 \text{ K}$))
α_{SG}	Amount of surface grains
γ_{i}	Shear of the i th slip system (MPa)
ε	Strain
$\dot{\varepsilon}$	Strain rate (s^{-1})
ε_{fr}	Elongation after fracture
ε_{g}	Elongation at maximum load
θ	Volume fraction of the border grains
$\Phi_{\%}$	Thickness reduction due to cold forming (%)
ρ_{dis}	Dislocation density ($1/\text{m}^3$)
ρ_{gd}	Gliding dislocation density ($1/\text{m}^3$)
σ	Normal stress (MPa)
σ_0	Scale parameter of the endurance limit
σ_{a}	Stress amplitude
σ_{e}	endurance limit
σ_{e0}	endurance limit of a smooth specimen
σ_{en}	endurance limit of a notched specimen
σ_{max}	Maximum stress
σ_{M}	Ultimate tensile strength (MPa)
σ_{N}	Apparent stress (MPa)

$\sigma_{N,en}$	Nominal endurance limit of a notched specimen (MPa)
σ_{pl}	Yield strength/elastic limit (MPa)
$\sigma_{pl0.2}$	0.2 %-yield strength (MPa)
τ	Transverse strain (MPa)
τ_N	Overall shear (MPa)
τ_{SC}	Single crystal critical resolved shear stress (MPa)
φ	Logarithmic degree of deformation [= $\ln (l/l_0)$]

3.1 Development of Microstructure

Martin Hunkel

3.1.1 Microstructural Size Effects

The plastic behavior of metals depends strongly on microstructural features such as dislocations, grain boundaries or precipitations. Some of these microstructural features show size effects depending not only on the size of the feature, but also on the size of the specimen. The influence of the microstructure on size effects can be roughly classified into four cases [11]:

- Microstructural or intrinsic size effects.
Dislocations interact with intrinsic crystallographic defects, like grain boundaries or precipitations.
- Statistical size effects.
In small sized samples, only a few grains exist across the thickness. Thus some single grain orientations can influence the whole plastic deformation.
- Lattice curvature and strain gradients.
Geometrically necessary dislocations are needed to compensate lattice curvature (bending) very strongly in thin sized specimens.
- Surface and interfacial constraints.
The surfaces themselves have an influence on grain boundary orientation and surface grain deformation, and interact with dislocations.

Besides the influence of the microstructure on size effects, the development of the microstructure (grains, dislocations) is size dependent in the case of interactions with the surfaces.

3.1.2 Grain Structure

In general, sheets have to be recrystallized after rolling for further forming processes due to high work hardening which comes with a high dislocation density. The

recrystallization is related to the change in grain size, grain shape, and texture. In the first stage, recovery and primary recrystallization occur in the same way as in bulk materials. Afterwards, normal grain growth can differ due to the free surface. Secondary recrystallization occurs in the presence of dissolving precipitations. In thin sheets, due to orientation effects, additional tertiary recrystallization (often also referred to as secondary recrystallization) leads to a pronounced texture.

If the typical length of the microstructure is very small compared to the size of the smallest dimension of a sheet or a wire, the behavior can be described as bulk. But if the length of the microstructure is comparable to the smallest dimension of a part, size effects occur due to the microstructure. For example, in the case of grains in sheets, the sheet can be described as bulk as long as the grain size d_G is small compared to the sheet thickness s . Hence, for a given sheet thickness, the smaller are the grains, the smaller are the size effects.

3.1.2.1 Grain Structures in Thin Sheets

Often, the surface grains differ in strength and shape from interior grains. For this purpose, the amount of surface grains of thin sheets is given by

$$\alpha_{SG} = \frac{2d_G}{s} \quad (3.1)$$

It is obvious that nearly all grains are surface grains if $d_G = s/2$. With this definition, a material can be described as bulk if $\alpha_{SG} \approx 0$.

Figure 3.1 shows micrographs of two sheets of austenitic steel X5CrNi18-10 (SAE 304) with thicknesses of 20 and 50 μm , respectively, and almost similar grain sizes. While in the 20 μm sheet there are only two to three grains across the thickness and $\alpha_{SG} = 0.8$, in the 50 μm sheet about ten grains exist across the sheet thickness and $\alpha_{SG} = 0.3$.

In cases of very large grains (much larger than the sheet thickness), they have a more two-dimensional grain structure with columnar grains (Tiffany structure, see

Fig. 3.1 Grain boundary structures of a 20 μm (*top*) and a 50 μm (*bottom*) thick steel sheet of X5CrNi18-10 (SAE 304)

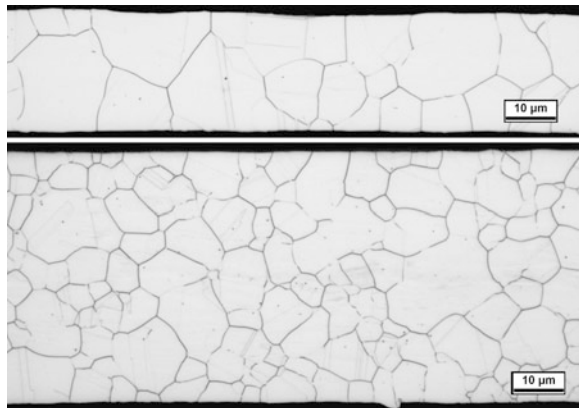


Fig. 3.2 Two dimensional grain structure within a sheet (Tiffany structure)

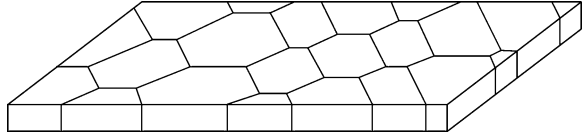


Fig. 3.2). Grain boundaries are all perpendicular to the free surfaces. Naturally, in a two-dimensional grain structure, the grain shape is anisotropic: The grain size in the sheet thickness is much smaller than the grain size within the plane of the sheet. Consequently, the grain size d_G within the plane of the sheet should be defined. Similar to a Tiffany structure in sheets, in wires there is a one-dimensional grain structure called the Bamboo structure. In a Bamboo structure no intersection between three grains exists. Finally, the extreme case is a single crystal where no grain boundary exists.

3.1.2.2 Normal Grain Growth

After recovery and primary recrystallization, a fine grained material with a high amount of grain boundaries exists. The driving force of normal grain growth is the reduction in the grain boundary energy, which is equivalent to the reduction in the grain boundary area. In bulk materials, the mean grain size d_G , dependent on time and temperature, can be described by

$$d^m - d_0^m = c_0 e^{Q/RT} \cdot t \quad (3.2)$$

with the growth exponent $m \approx 2$. Typically, the grain sizes during normal grain growth are distributed log-normally. The grain growth velocity lowers significantly when the grain size reaches the size of the sheet thickness. In the case where $\alpha_{SG} \rightarrow 1$, the characteristic of the grain growth changes from three-dimensional to two-dimensional. While for bulk material the grain boundaries move in all three directions, the further motion of the grain boundaries in thin films lies in the plane of the sheet. This reduced dimensionality reduces the growth velocity significantly [49]. Additionally, thermal etching leads to grooves at the intersection of the grains with the surface. Especially, the growth exponent m can be significantly larger in thin films than in bulk materials (m up to 10 [14]).

In single phase alloys, the equilibrium condition leads to an angle of 120° at the edges of three grains and 109° at the corners of four grains. Due to this difference in angles at the edges and corners, grain growth always occurs in bulk material. If a grain boundary encounters a flat surface, the preferred angle is 90° . For the Tiffany structure, shown in Fig. 3.2, no corners with four grains exist. All grain boundary angles are 120° . Thus there is almost no driving force for further grain growth. Of course, the grain boundary energy has not attained its absolute minimum (grain boundaries still exist) but locally the driving force is very small and other effects have to contribute to a driving force for further grain growth. Such an effect is the orientation-related grain growth (tertiary recrystallization) described below.

Fig. 3.3 Grain boundary encountering a free surface without (*left*) and with (*right*) thermal etching



It is well-known that grain boundaries can be pinned very effectively at precipitations. At a free surface another pinning effect can occur. Recrystallization takes place at higher temperatures ($T > 0.4 T_M$). Therefore, thermal etching can lead to grooves, see Fig. 3.3. These grooves pin the grain boundaries very effectively and, therefore, the grain growth reduces significantly [39].

3.1.2.3 Orientation-Related Recrystallization (Tertiary Recrystallization)

Sometimes, after a stage of normal grain growth, abnormal grain growth occurs. During abnormal grain growth a few grains grow significantly faster than the other grains. The resulting distribution of the grain sizes is no longer a log-normal distribution but a bimodal distribution. In some materials, a pronounced texture is found due to abnormal grain growth. This orientation-related recrystallization is an important feature in thin iron–silicon- (Fe–Si-) sheets for electrical applications [52], where texture has a strong influence on magnetic properties. In thin Fe-3 % Si films, orientation-related recrystallization leads to a pronounced and very sharp (110) [100] Goss texture.

The underlying reason for orientation-related grain growth is still under debate. Theories exist for orientation-related recrystallization, such as selective nucleation or selective grain growth [55]. The stagnation of normal grain growth seems to be a requirement. As described above, in Tiffany-like structures the driving force for further grain growth due to grain boundary reduction is very small. Another possibility for the stagnation of further grain growth is the pinning of grain boundaries at precipitations or grooves. In the case of the stagnation of normal grain growth, other energy reduction effects can lead to further grain growth, which is suppressed due to the much smaller energy reward during normal grain growth. Reasons for the selective growth of grains with specific orientations have been discussed. The local energy of a free surface depends strongly on the crystallographic orientation of the grain with respect to the free surface. Grains with a preferred orientation can grow at the expense of less preferred grains. Another theory is that grain boundaries with a misorientation angle of 20–45° have high mobility. Goss grains often have such boundaries and the growth of these grains is preferable.

3.1.3 Dislocations

The deformation of parts during forming is due to the generation and movement of dislocations. The movement of the dislocations is restricted to active glide

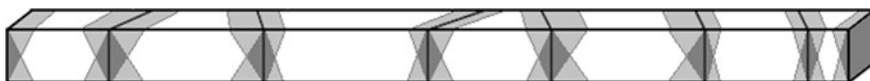


Fig. 3.4 Softer (*white*) and harder (*grey*) regions within a sheet with a Tiffany structure

systems. In bulk materials, among others, grain boundaries act as a barrier for further dislocation movement. The piling-up of dislocations at the grain boundaries leads to a yield strength increase proportionally inverse to the square root of the grain diameter (Hall-Petch relation).

3.1.3.1 Dislocations at Free Surfaces

The relation changes in the presence of a surface and especially according to the nature of the surface. There are two extremes in the nature of a surface. A layer on the surface, like an oxide-layer, can block the dislocations. On the other hand, a totally free surface can act as a source or sink for dislocations. In thin sheets not only does pile up of the dislocations at grain boundaries occur, but also pile up of dislocations at passivated surfaces. Consequently, sheets with passivated surfaces have a higher work hardening effect than sheets with unpassivated surfaces [54]. Grains with free surfaces consist of a softer central region and a harder region near grain boundaries (Fig. 3.4). In the softer inner region, the dislocations can move on the active slip systems to the surfaces and disappear if the surface is unpassivated [20]. Therefore, this region has a significantly lower dislocation density and is significantly softer than the core of the grain in bulk materials. Dislocations near grain boundaries pile up at the grain boundaries and these regions are harder.

Typically during plastic deformation, dislocations pile up at grain boundaries as well as at passivated surfaces, and Orowan loops occur around precipitations. After applying a reverse loading, these dislocations can move easily in the reverse direction and significant lower yield strength occurs than without a pre-straining. This effect is called the Bauschinger effect. If layered surfaces act as a barrier for dislocations, then they pile up at the surface during loading. Afterwards, during unloading, a pronounced Bauschinger effect occurs. On the other hand, for unlayered surfaces, the dislocations leave the material and a lesser Bauschinger effect occurs.

3.1.3.2 Geometrically Necessary Dislocations

Single crystals under uniaxial stress can deform without the generation of inner dislocations. For torsion or bending, plastic strain gradients exist. These plastic strain gradients have to be compensated by additional dislocations, generally called geometrically necessary dislocations (GNDs). Also, plastic strain gradients are needed to compensate the different strain behaviors of grains in polycrystals. GNDs lead like

other dislocations to an additional hardening effect. Furthermore, the GND field is strongly dependent on the size of the specimens, as the plastic inhomogeneities between the grain boundaries and the cores have to be accommodated by the GND field within a varying distance for the different sample sizes. Therefore, thin sheets have higher hardening during bending than thick sheets [44, 46].

3.2 Heat Treatment

Roland von Bargen, Axel von Hehl and Hans-Werner Zoch

3.2.1 Annealing Processes

During cold forming the formability of the metal decreases with increasing forming ratio due to the rising number of dislocations that hinder further plastic deformation. On the one hand, the local stress fields of the dislocations interact and increase the macroscopic stress, which has to be induced to achieve further deformation. On the other hand, the dislocations accumulate at the grain boundaries and by this the free run length on the deformation planes that can be used by following dislocations decreases [8].

To regain formability for further cold forming operations, heat treatment processes like recrystallization annealing can be performed. This works for steel as well as for aluminum alloys. Different recrystallization mechanisms can be distinguished: primary recrystallization, strain induced grain boundary movement, general grain growth and secondary (ternary) recrystallization [8].

Primary recrystallization requires a sufficient amount of cold forming (critical forming ratio) before heating above the recrystallization temperature T_R . An approximation of this temperature can be calculated based on the melting temperature T_M in Kelvin using the formula:

$$T_R = 0.4 \cdot T_M \quad (3.3)$$

During recrystallization, new grains start growing homogeneously from preferred nucleation sites, like triple points of the former grain boundaries. The new grains grow further, mostly globularly shaped, and consume the old anisotropically deformed ones. Parallel to this the dislocations generated by cold forming are eliminated by annihilation due to diffusion-driven atom movement at elevated temperatures. The energy that was consumed by the material during forming is the driving force for the grain growth and dislocation movement [8]. Primary recrystallization is a very fast process compared to the others where only grain growth is occurring. The resulting grain size mainly depends on the deformation

degree. With increasing forming ratio the resulting new grain size decreases. Other process parameters like recrystallization temperature, the heating rate or the grain size before forming have a negligible effect compared to the deformation degree [8]. The critical deformation degree is different for each material. Soft steels for example require 5–10 % deformation, aluminum alloys sometimes only 2 up to 15 % [23].

If the deformation degree is not sufficient for primary recrystallization the existing grain boundaries can move, local grain growth will occur and the effects induced during forming will be reversed (e.g. by dislocation annihilation). The grain growth will start at grains with lower local deformation and grains that saw higher deformation will be consumed. Therefore this mechanism is called strain induced grain boundary movement [8].

After the new grain structure has been developed by one of the above-mentioned mechanisms further grain growth can occur. Because the energy induced by the plastic deformation is already utilized, other driving forces like the grain boundary or the surface energy are necessary. Compared to the deformation energy these forces are very low, but the grain boundary energy in some cases is sufficient to enable the ongoing movement of the grain boundaries. If many grains increase nearly equally in size while consuming only some grains the mechanism is called general grain growth. In some cases only a small amount of grains increase (significantly) in size, which is called secondary recrystallization. For surface grains another mechanism called ternary recrystallization can occur. The driving force is the surface energy of the grains. If a grain has a lower surface energy compared to its neighbors because of its orientation, it will grow and consume its surrounding grains with the higher surface energy. This effect often occurs within thin sheets where there is only one huge grain stretching over the material's thickness [13]. Focusing on very thin micro metallic components the latter mentioned mechanism could result in inhomogeneous grain structures with anisotropic or even very locally controlled material properties. Figure 3.5 gives an overview of some of these mechanisms in correlation to the degree of deformation.

A typical microstructure resulting after recrystallization annealing is shown in Fig. 3.6. In this case mainly primary recrystallization took place. The short term annealing process leads to a very fine and homogeneous grain structure due to the very high reduction in diameter by prior cold forming.

3.2.2 Final Heat Treatment

3.2.2.1 Martensitic Hardening of Steels

Martensitic hardening is the most important and most effective strengthening mechanism for hardenable steels. These steels undergo a required phase transformation from ferrite/pearlite to austenite during heating from room temperature up to the austenitization temperature, which is around 800–950 °C depending on

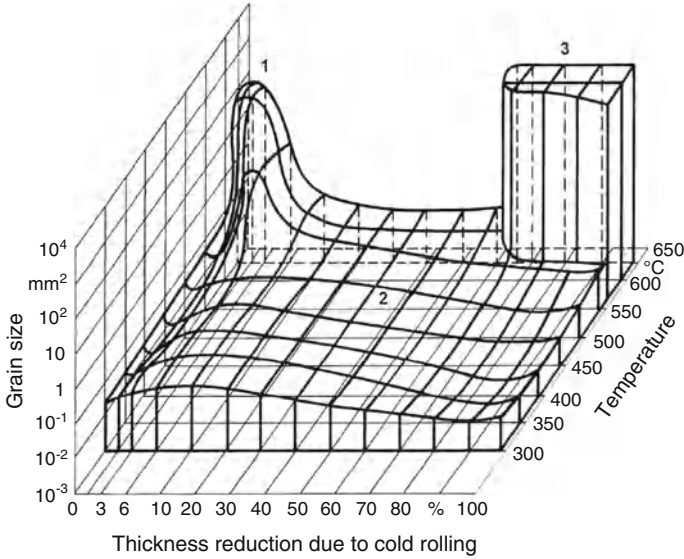


Fig. 3.5 Recrystallization diagram: grain size of EN AW-Al99.6 (EN AW-1060) depending on thickness reduction and recrystallization temperature, annealing time 2 h; 1 coarse grains due to primary recrystallization near the critical deformation degree, 2 area of small and medium grains due to primary recrystallization, 3 coarse grains due to secondary recrystallization according to Kammer [23]

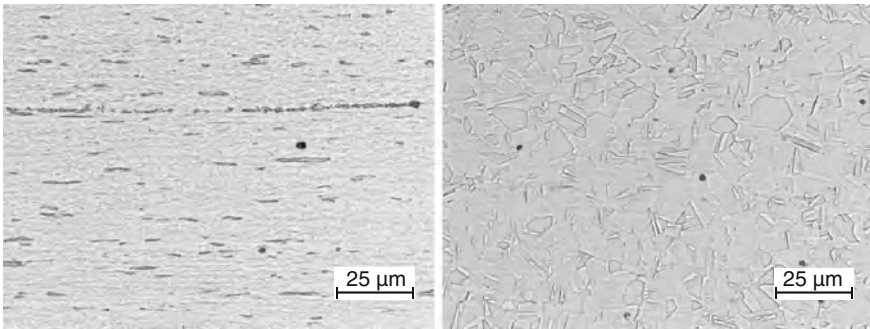


Fig. 3.6 X5CrNi18-10 steel wire: *left* rotary swaged from Ø 2 mm down to Ø 0.4 mm, *right* recrystallization annealed in a drop-down tube furnace at 1,300 °C for 3.4 s

the steel composition. Therefore austenite stabilized corrosion resistant steels like X5CrNi18-10 (SAE 304) that show no phase transformation between the temperatures mentioned cannot be hardened by forming martensite. The holding time at austenitization temperature has to be sufficiently long to bring all the carbon out of carbides in solution that can be dissolved in the austenite lattice at that

temperature [8]. Furthermore a homogeneous distribution of the alloying elements should be accomplished. Information about necessary austenitization times can be taken from time temperature austenitization diagrams (TTA). For micro components the heat treatment times should ideally be short to match the high production rate of the components. Special furnace concepts can be used, which are described in Sect. 3.2.3.3, but allow only very short austenitization times. Therefore the homogeneity of the base material has a crucial influence on the resulting material properties because the higher it is the shorter the duration to achieve a homogeneous austenite. During cooling different phases, like carbides, ferrite, pearlite, bainite, can be formed by diffusion processes depending on the cooling rate [8]. The resulting composition and hardness can be estimated for example with continuous-cooling transformation (CCT) diagrams that are available for most steels from databases. If the material is quenched (cooled very fast) no diffusion based phase transformation is possible and diffusion-less martensitic transformation takes place. During this transformation the face-centered cubic austenite lattice shears into the tetragonal body-centered martensite lattice [8], as Fig. 3.7 illustrates. For micro components the cooling rate generally is very high due to the small volume to surface ratio. Therefore the quenching media, for example ambient air or water at room temperature-both without agitation-has no significant influence on the resulting microstructure and its mechanical properties as shown by Fig. 3.8. Depending on the austenite grain size of the micro component, a martensitic microstructure will develop. Small austenite grains will lead to a martensitic microstructure consisting of fine needles, lathes or plates. Whereas bigger grains enable larger areas with the same crystallographic orientation to shear at once, which can result in a coarser martensitic microstructure.

The resulting martensitic microstructure of the steel after hardening is significantly influenced by the carbon content. Regarding the martensite start (start of the austenite into martensite transformation) and the martensite finish (complete transformation of austenite into martensite) temperature, Fig. 3.9 shows that with an increasing carbon content both temperatures are shifted to lower temperatures. Steel with 0.5 mass % carbon reaches the martensite finish temperature at approximately 20 °C. For higher carbon contents the austenite into martensite transformation at room temperature is incomplete (Fig. 3.9) and retained austenite remains [8]. This retained austenite can possibly transform into martensite under the influence of strain or temperature, for example when the component is in use. The hardness of the martensitic microstructure increases first with increasing carbon content (Fig. 3.9). At some point an increase in carbon content after full austenitization results in a decrease in hardness, which is due to the higher retained austenite volume fraction with its lower hardness. Therefore for this type of steels the hardening temperature is mainly chosen in the two-phase region austenite-carbide to prevent excessive retained austenite [8]. Figure 3.10 illustrates the two main types of martensitic microstructure: lath and plate-shaped martensite in relation to the carbon content of some exemplary steels.

Fig. 3.7 Transformation from austenite (*top*) to martensite (*center*) by lattice shearing according to Macherauch and Zoch [32]

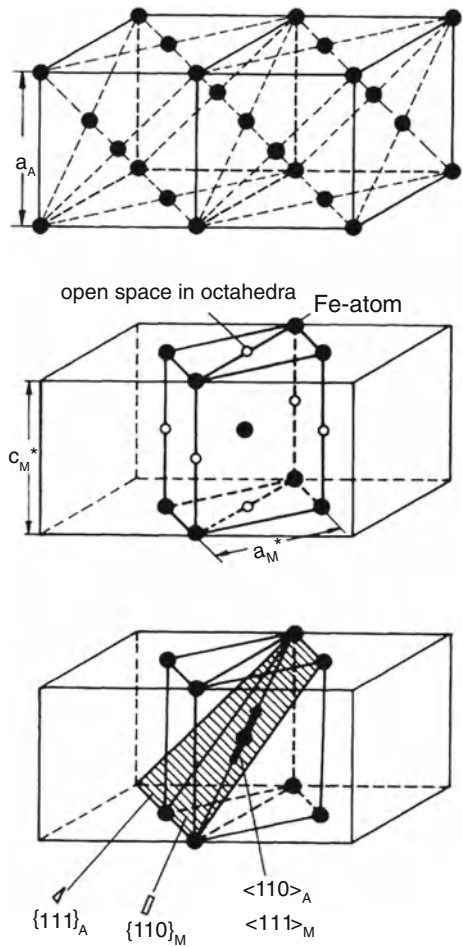
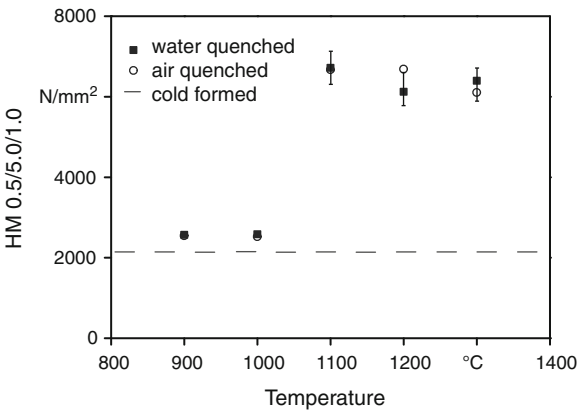


Fig. 3.8 Hardness of micro cold-formed cups (diameter 4 mm, sheet thickness 50 μm) made out of plain carbon steel C100 and quench hardened in a drop-down tube furnace with different quenching media



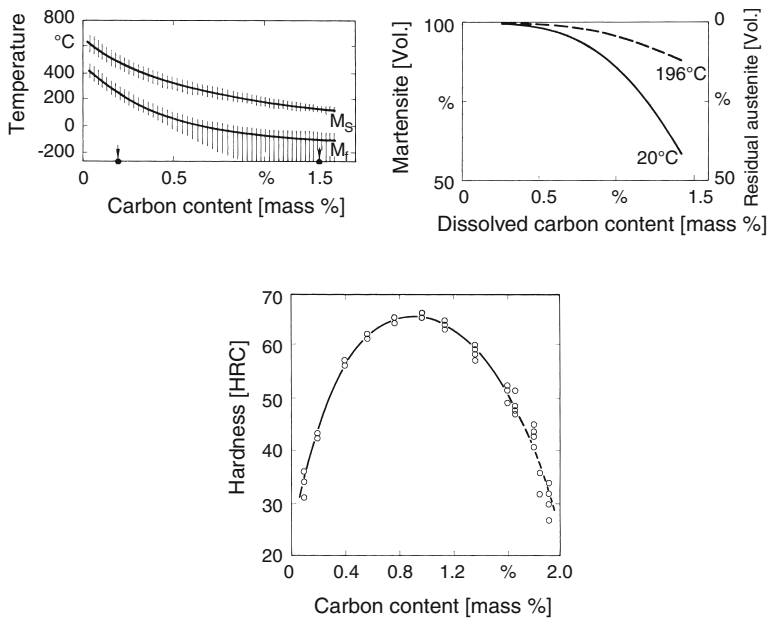


Fig. 3.9 Influence of carbon content on the martensite start / finish temperatures, the retained austenite volume fraction and the resulting hardness according to Macherauch and Zoch [32]

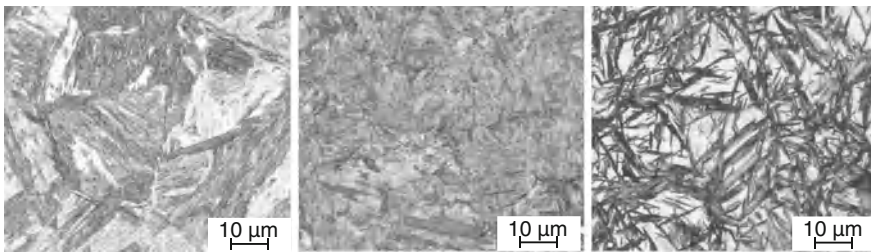


Fig. 3.10 Martensitic microstructure of unalloyed steels (*left to right*): lath martensite (Ck 15), mixed martensite (Cf 53), plate martensite (125 Cr 2) according to Macherauch and Zoch [32]

3.2.2.2 Precipitation Hardening of Aluminum Alloys

Aluminum alloys are generally divided into two groups: wrought and cast alloys. Each group is sub-divided into alloys that can only be strengthened by cold forming and age hardenable alloys, which can be significantly strengthened by heat treatment. An alloy that is age hardenable has to fulfill three main prerequisites [42]:

1. Decreasing solubility of the alloying elements in the matrix lattice with decreasing temperature.
2. Existence of an intermetallic phase at room temperature in the phase diagram.
3. Vacant lattice positions that allow diffusion processes at room or elevated temperature.

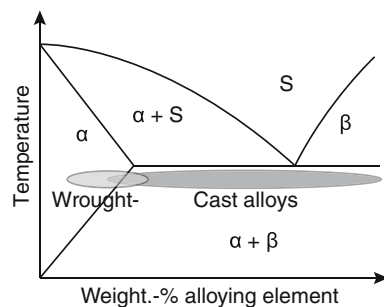
The heat treatment process of age hardening, or in other words precipitation hardening, consists of three major steps: solution annealing, quenching and aging.

Solution annealing is performed at temperatures around 470–560 °C depending on the specific alloy. The goal is to dissolve the alloying elements atoms in the aluminum α matrix lattice [42]. For hypoeutectic alloys, which are most of the wrought alloys, (nearly) all alloying elements can be dissolved. Hypereutectic alloys, like most cast alloys, only allow the eutectic concentration of the alloying elements atoms to be dissolved, see Fig. 3.11.

Another important effect during solution annealing is the increasing vacancy density due to the high temperature. Regarding process control it is important to keep the solution annealing temperature below the eutectic temperature, otherwise partial melting of segregated areas with higher alloying element saturation is possible. These areas are most often located at the grain boundaries [42]. If partial melting occurs a significant decrease in the structural strength will be the consequence, most likely making the component unusable.

After the component is solution annealed it has to be quenched to room temperature. During quenching it is important to cool the component as fast as necessary to force all the dissolved alloying element atoms to remain in solid solution, but as slowly as possible to avoid distortion of the component. The material state after quenching is called supersaturated solid solution, which is an unstable non-equilibrium condition. If the quenching is not fast enough, intermetallic precipitates of the equilibrium phase will form that have no strengthening effect but bind alloying element atoms, which are no longer available for hardening precipitates in subsequent aging treatments [42]. Furthermore if the quenching is fast enough the high vacancy density can be maintained, which facilitates diffusion processes. Quenching of aluminum alloys is in most cases carried out with water or polymer solutions. Alloys that do not require fast quenching to maintain the supersaturated solid solution can be quenched by (high pressure) gas quenching or by liquid–gas sprays.

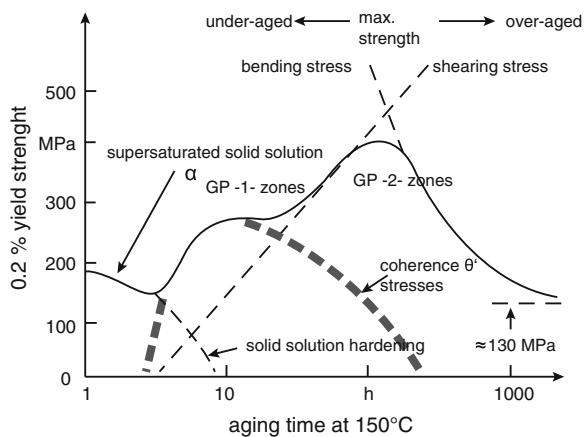
Fig. 3.11 Schematic eutectic phase diagram of aluminum alloys and typical solution annealing temperature range for wrought and cast alloys (α aluminum matrix, S melt, β intermetallic phase of the specific alloying system)



The final step of the age hardening process is the aging treatment. Aging can be performed as natural aging at room temperature or as so-called artificial aging at elevated temperatures between 120 and 200 °C. During aging the dissolved alloying elements precipitate along one or more precipitation sequences that are specific for each alloying system [42]. First of all the alloying element atoms begin to cluster in the matrix lattice. These clusters become more and more organized and unstable coherent precipitates form, for most alloys called Guinier–Preston-zones (GP-zones). Depending on the alloy these GP-zones are sometimes separated in GP-I- and GP-II-zones, which have a higher grade of organization and complexity. By natural aging only these precipitates can be formed. Therefore the increase in hardness is mainly caused by coherency strain. During an artificial aging treatment the GP-zones dissolve or transform into metastable semi-coherent precipitates that are precursors of the equilibrium phase. Finally the stable incoherent equilibrium phase of the precipitation sequence forms. Sometimes an alloying system allows more than one precipitation sequence during artificial aging treatments. The equilibrium phase and sometimes its precursor cause a decrease in material strength, which is called overaging. Generally the precipitate size and the mean distance between them increase with longer periods of aging [42]. The strengthening effect of an aged material mainly depends on two mechanisms that dislocations can use to pass precipitates. Small coherent precipitates can be sheared, described as the Kelly-Fine mechanism. Larger, incoherent ones are passed by the bending of the dislocation, which is described by the Orowan mechanism, leaving behind dislocation circles around the particles. The maximum hardness respective to the strength of the material is reached, if both mechanisms are equally likely to be used by dislocations as Fig. 3.12 indicates. At this point the optimal precipitate diameter and mean distance is obtained [42].

For most alloys, the maximum hardness reached by artificial aging will surpass the one after natural aging. Furthermore the process time can be significantly reduced by artificial aging due to the higher diffusion speed. Nevertheless it has to be mentioned that the higher the artificial aging temperature is, the faster the

Fig. 3.12 Development of strength during artificial aging of the aluminum alloy EN AW-AlCu4 according to Keßler [24]



maximum strength that is reached, but at the same time the maximum strength is reduced [42].

As well as for martensitic hardening of steel, the heat treatment times for micro components made from aluminum should be as short as possible to match the fast production rates of the cold forming processes. Heat treatment facilities, like the drop-down tube furnace (Sect. 3.2.3.3) enable short processes, but require special material characteristics. Conventional aluminum alloys could hardly be precipitation hardened in such a facility, because the solution annealing required as well as the aging times are too long for most of them. For example, the drop duration of an aluminum cup in the drop-down tube furnace is around 7 s. A typical high strength wrought alloy Al-4.5Zn-1 Mg (EN AW-7020) has a solution annealing time of around 20 min, which is two orders of magnitude higher. Therefore alloys that are suitable for age hardening in for example a drop-down tube furnace should already be in solution-annealed condition after metallurgical fabrication. This can be achieved by PVD processes (Sect. 6.1) or to some extent by spray forming, which allows faster cooling rates than casting processes.

3.2.3 Furnace Concepts

3.2.3.1 Quenching Methods: Liquid Quenching

Liquid quenching is in most cases the fastest technique to cool metallic components from elevated temperatures, because the highest heat transfer rates can be achieved [31]. Therefore steel parts are very often quenched in oil and aluminum parts in water or water-based polymer solutions. The main differences between the liquid quenching media mentioned are their boiling points, with oils having the highest, water the lowest and polymer solutions lying in-between. A huge issue when quenching with liquid media is the Leidenfrost phenomena [31]. This describes the effect that occurs when a hot component is dipped into a quenching bath with its temperature above the fluid's boiling point. In this case the fluid evaporates immediately and forms a closed vapor film around the component. The heat transfer coefficient is very low in this phase as the vapor film insulates the part from the liquid. At edges or thin cross-sections of the component running ahead in cooling, the vapor film starts to break up and the nucleate boiling phase begins in these areas, hereafter covering the entire surface. In this phase the heat transfer is several times higher than in the vapor film phase, and the component cools rapidly. When the temperature has dropped below the boiling temperature only convection cooling takes place and the heat transfer coefficient decreases again, but is still higher than in the vapor film phase [31]. All together the cooling of a complex component in an evaporable liquid is always inhomogeneous due to the vapor film that breaks up locally. This results in the distortion of the component, or even damage like microcracks is possible.

Table 3.1 Liquid and gas quenching media and their heat transfer coefficients according to Köhler et al. [31]

Quenching media	Circulation	Heat transfer coefficient [W/(k × m ²)]
Nitrogen (at 1,000 hPa)	Yes	100–150
Nitrogen (at 5,000 hPa)	Yes, strong	300–400
Oil (20–80 °C)	No	1,000–1,500
Oil (20–80 °C)	Yes	1,800–2,200
Water (15–25 °C)	No	3,000–3,500

Another disadvantage of some liquid quenching media, like oils or polymer solutions, is that costly cleaning treatment has to be performed after quenching to remove the oil or polymer residues. Both disadvantages mentioned have even a greater impact on very small components, because distortion is harder to correct and cleaning is much more difficult due to capillary effects.

3.2.3.2 Quenching Methods: Gas Quenching

Gas quenching offers a great potential to overcome the disadvantages of liquid quenching media mentioned. Unfortunately the achievable heat transfer coefficient is much lower than for liquid quenching media as Table 3.1 indicates.

Advantages however are very homogeneous quenching in all positions of the component (no Leidenfrost phenomena can occur) and the fact that no cleaning step is necessary. Gas quenching is in most cases performed as a high-pressure gas quenching treatment, where the components are cooled down in the heating chamber or are transferred into a cooling chamber, and in both cases purged with a gas stream. Generally gases are used that are inert to the component material, like nitrogen or helium [31].

Very small components also have the benefit that their surface to volume ratio is in most cases fairly high, which makes gas quenching even more effective.

A special solution for these small components can be a cone-shaped quench chamber in which the components drop, are caught in a reverse gas stream and cool down while tumbling in the chamber [45]. Due to the cone-shaped chamber the gas velocity decreases from the bottom to the top. Therefore the components cannot be thrown out again or hit the bottom of the chamber.

3.2.3.3 Furnace Concepts: Drop-Down Tube Furnace

One of the challenges for micro components is the method of handling and charging the small and sensitive components. The conventional ways of charging for example on racks is unsuitable for small micro components because of the size and the high mounting effort. Baskets—generally used for components with a size of some millimeters—are also inappropriate, because an extremely tight mesh

would be necessary to hold the micro components. Besides, due to the intense mechanical interaction between the sensitive parts at high temperatures the deformation probability would be very high. Furthermore, as a result of contact, diffusion as a threat of welding effects for the components would also be high. To avoid these effects a new heat treatment process was developed by which micro components are dropped through a vertical tube furnace and for example are austenitized and finally hardened by quenching, leaving the tube at the bottom. To realize the idea of a heat treatment during falling in an applicable manner, an adequate facility is necessary. Requirements are a great height to have a long drop duration which correlates for example with the austenitizing time and a high maximum heating temperature in order to realize excess temperatures and accelerate the desired structural effects like the dissolution of precipitates or grain recrystallization. An example of a heat treatment device fulfilling both requirements is described below. It is a 6.5 m tall vertical drop-down tube furnace with a tube length of 5.5 m in total and an electric heating power of 34 kW (Fig. 3.13). From the one end to the other the furnace is divided into six separately controllable

Fig. 3.13 Drop-down tube furnace for heat treatment during falling



heating zones with a maximum temperature of 1,300 °C and a total heated length of 5 m. The inner tube diameter is 54 mm.

To feed in the components that should be heat treated, a component feed is necessary, which is one of the most critical and important devices of the drop-down tube furnace. It includes several functions that are pointed out below. For first experiences with such a device a manually working solution is suitable, allowing later automation to be integrated within the whole production chain. One of the main functions of the feeding device is to seal up the furnace at the upper end to avoid the stack effect that may hinder the components from falling due to the ascending hot atmosphere. The component feed designed is based on a retaining device for the components (Fig. 3.14). By turning the retaining device the component leaves the component feed and starts to drop along the central axis of the furnace. A diffusor mounted on top of the furnace enables rinsing with purge gas. The diffusor is necessary to achieve an almost laminar flow in the tube furnace that keeps the sensitive components away from the tube's inner surface. Furthermore, it is necessary to use an inert gas in order to avoid the destruction of the components due to oxidation. It has to be mentioned that molecular nitrogen dissociates above 1,050 °C and would react with steel components. In combination with hydrogen the critical temperature is reduced to 750 °C [9].

The gas flow characteristics can be varied in a wide area. For the furnace described with a volume of around 14 l a gas flow was realized with argon and an average velocity of 0.003 m/s, which correlates to a flow rate of 25 l/h. This was sufficient to avoid oxidation during heat treatment.

To achieve repeatable heat treatment results in conventional heat treatment facilities it is necessary to keep process times constant. Regarding the different shapes and flow characteristics of complex components this seems to be a critical requirement for processes in the drop-down tube furnace. Measuring the drop duration is therefore very important to ensure the comparability of results. One

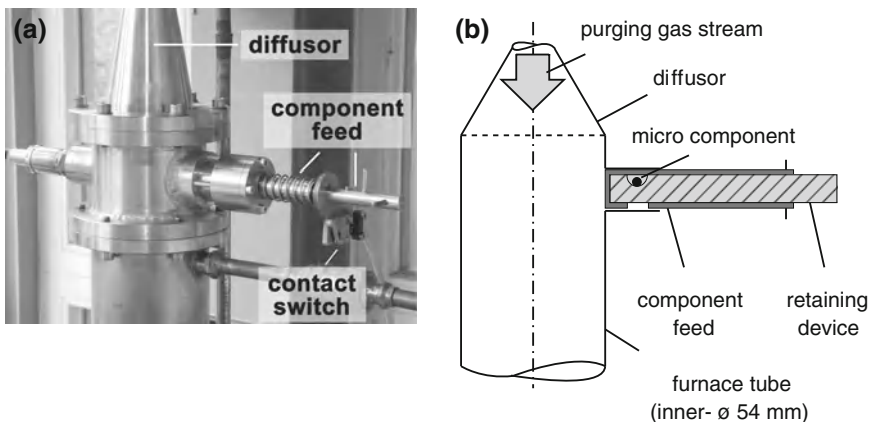
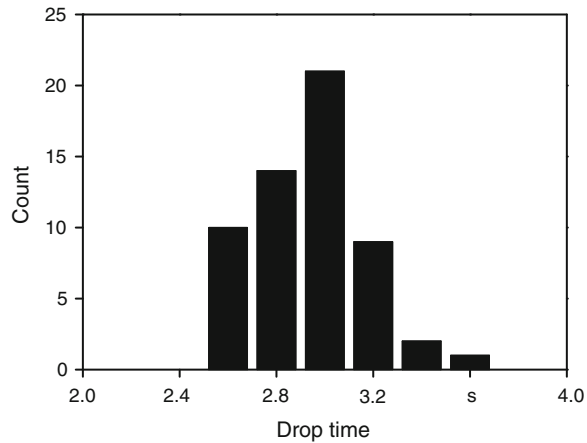


Fig. 3.14 **a** Component supply with attached diffusor. **b** Schematic drawing

Fig. 3.15 Drop duration measurements of C100 steel cups at 20 °C furnace temperature; geometry of the cups: *diameter* 1 mm, *height* 0.5 mm, *material thickness* 20 μm



possibility for measuring is a real laser light barriers that have a short pulse length, to detect the fast falling components, and which offers a high sensitivity regarding the difference between sent and incoming light that is caused by parts below 1 mm in diameter. Attached to the described drop-down tube furnace is a measuring system, which is activated by a contact switch, integrated in the component feed when the component drops out of the retaining device. At the bottom of the furnace the detection of the components is carried out by the areal laser light barriers mentioned, with a pulse length of 3.5 μs , which cover the complete width of the tube and can detect objects down to 0.3 mm in diameter.

Drop duration measurements of steel cups with a diameter of 1 mm show that the effect of different possible orientations during falling only influences the heat treatment time by $\pm 14\%$ (Fig. 3.15).

3.2.3.4 Furnace Concepts: Belt Furnace and Pusher Tray Furnaces

Belt furnaces are often used for components within a size range of 1–10 mm in all three spatial directions, like bolts, nuts, springs, balls, rollers or chain segments [30]. The components are placed as bulk material on a heat resistant conveyor belt and then move slowly through the horizontal furnace. The furnace itself is purged with an inert protective gas to avoid the oxidation of the components [5]. Sometimes the components drop directly in a quenching bath at the exit of the furnace. For components smaller than 1 mm or thin foils below 100 μm in thickness this furnace concept is inadequate due to several reasons already mentioned in Sect. 3.2.3.3.

Pusher tray furnaces also often are used for smaller bulk components, similar to belt furnaces. However the homogeneity of heating and prevention of distortion make this furnace type unsuitable for micro components.

3.3 Flow Stress

Bernd Köhler and Hubert Bomas

This section is headed “flow stress”, which is often used in connection with forming. Flow stress means the stress along one axis at a given value of strain that is in general required to produce plastic deformation. In contrast with this, in this section the term “yield strength” is used simultaneously. Yield strength means “the engineering stress at which, by convention, it is considered that plastic elongation of the material has commenced” [35]. This stress is usually specified in terms of a specified deviation from the linear stress–strain relationship. For example, the common 0.2 % yield strength $\sigma_{p0.2}$ is defined as the stress where 0.2 % plastic strain occurs.

3.3.1 Tensile and Compression Tests

3.3.1.1 Tensile Tests

Tensile tests are used to characterize the elastic properties and the resistance against plastic deformation, as well as against fracture at positive deformation rates. In this, the specimen will be drawn at a constant strain rate until failure occurs, while the required force is measured. The dog bone-design of a tension test flat specimen with a thickness of less than 1 mm is in principal aligned on the ISO 6892-1 standard for tensile testing of metallic materials, and is shown in Fig. 3.16a. To ensure that failure takes place within the gauge length, the heads of the specimen are broader than the central area and the transition is shaped as a radius. Moreover, at the testing of very thin sheets special attention has to be given to the quality of the sample’s edges, which have to be smooth and without micro notches resulting from the cutting process. Another important parameter for the tensile test of a thin sheet is to minimize the eccentricity of the specimen in the driving direction, which if it is too big may cause wrinkling and early failure [18]. To avoid these failures it is helpful to use an appropriate mechanical adjustment instrument. Generally, the handling of samples with thicknesses in the sub-millimeter range requires the utmost care, using adequate equipment like special forceps such as vacuum tweezers. It is also possible to protect mechanically sensitive specimens by using special carriers or support frames, which are not removed until the specimen is installed in the testing machine [18].

Figure 3.16b shows an engineering stress–strain diagram, which means a plot of the apparent stress $\sigma_N = F/A_0$ versus the strain ε , at which F is the currently required force and A_0 is the specimen’s cross-sectional area at the beginning of the test. At the start of deformation, between points A and B, the material behaves completely elastically obeying Hooke’s law ($\sigma_N = E \cdot \varepsilon$), in which the elastic

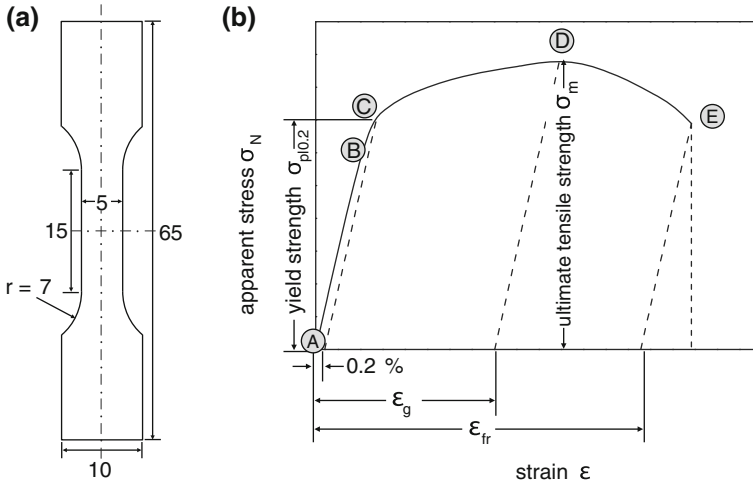


Fig. 3.16 Tensile flat specimen and schematic stress–strain diagram (not to scale)

behavior is characterized by the slope E of the straight, called Young's modulus. At point B the curve deviates from linear behavior with respect to the start of plastic deformation. The corresponding stress σ_{pl} is called the elastic limit, which is of great importance for all kinds of forming processes. This stress is difficult to determine, since at first plastic behavior takes place gradually. To avoid the uncertainties of small plastic strains a yield strength measurement is carried out, usually measuring the stress at 0.2 % plastic strain (point C), called the 0.2 % yield strength $\sigma_{pl0.2}$. Afterwards the apparent stress σ_N constantly increases until point D, called the ultimate tensile strength σ_m where the maximum apparent stress is achieved. The plastic strain corresponding to σ_m is called the elongation at maximum load ϵ_g (Fig. 3.16b). With further deformation necking appears, which means a local reduction of the cross-section corresponding with a decrease in the apparent stress until failure occurs at point E. The appropriate plastic strain ϵ_{fr} is termed elongation after fracture.

3.3.1.2 Compression Tests

A second basic experiment with well-known stress and strain conditions is the uniaxial compression or upsetting test, according to DIN standard 50106 which covers the macro-range. Compression tests usually are performed between two coplanar, ground disks, which upset a cylindrical sample at a constant rate with the simultaneous measurement of the upsetting and required forces. The upset measurement can take place in a non-contact manner with capacitive or inductive displacement sensors, while the force is usually measured by a load cell. In the

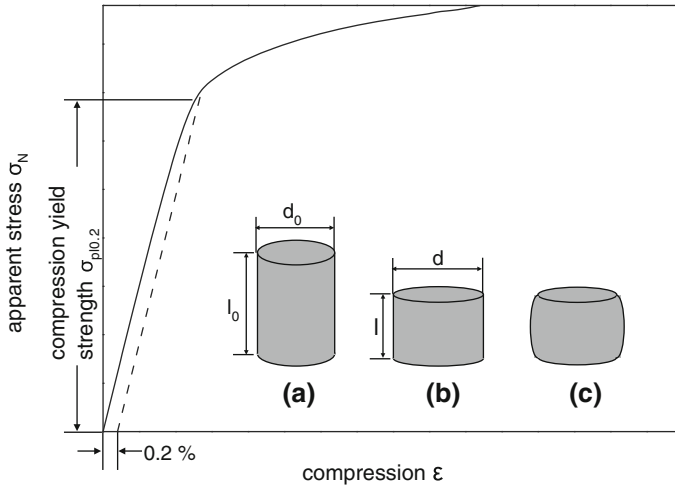


Fig. 3.17 Schematic stress–compression diagram (not to scale) and effect of lubrication on the upset test: **a** prior to the test, **b** optimal lubrication: specimen remains cylindrical, **c** insufficient lubrication: specimen obtains convex contours

micro range, standard cylindrical specimens exhibit a length to diameter ratio l_0/d_0 of 0.5 to 2, as in the macro range, however they are smaller. Typical diameters for micro upsetting specimens are within the range of 1–10 mm [36].

An important precondition for valid results is that the specimen remains cylindrical during the test. Otherwise it is nearly impossible to determine the logarithmic degree of deformation $\varphi = \ln(l/l_0)$ in the correct manner. To ensure such homogenous forming it is necessary to have proper lubrication between the dies and the front surfaces of the specimen (Fig. 3.17). This applies all the more the smaller the sample. Furthermore, parallelism deviation of the sample's front surfaces leads to an inhomogeneous stress–strain distribution and a flattening of the scope in the material's stress–strain response [16].

Figure 3.17 shows a schematic stress–compression diagram with the 0.2 % compression yield strength $\sigma_{pl0.2}$, which is determined along the lines of $\sigma_{pl0.2}$ in the tensile test. The apparent stress $\sigma_N = F/A_0$ is in reference to the specimen's front surface area A_0 at the beginning of the test. In fact, with respect to forming it is more adequate to use the flow stress $\sigma_f = F/A$ where $A = A_0 \cdot l_0/l = A_0/(1 + (l - l_0)/l_0)$ is the specimen's real front surface area during the experiment.

3.3.2 Classes of Size Effects Related to Yield Strength

With respect to the influence of size effects on the yield strength, three main areas can be identified:

The first and strongest influence is caused by mechanisms which inhibit dislocation movement. This effect typically occurs in specimen sizes between 1 and 10 μm . It is due to a lack of dislocation sources or drains or due to a surface layer, which suppresses the dislocation annihilation giving rise to very effective dislocation pile ups accompanying an increase in yield strength.

The second area concerns strain gradient plasticity effects. A corresponding model has been developed primarily from indentation hardness experiments. As a basic approach a strain gradient model differentiates between two types of dislocation, statistically stored dislocations (SSD) and geometrically necessary dislocations (GND). SSD are those dislocations which are created during plastic deformation and are statistically homogeneously distributed within the crystal. In contrast, GND are additionally incorporated dislocations to compensate for inherent strain gradients in the lattice and are distributed accordingly. The interaction width between a GND and mobile dislocations are described using length scales, which are first of all determined by the material but are also influenced by the respective underlying theory. Those length scales are reported to be in the order of magnitude between 10 nm and 100 μm . On the basis of these characteristic length scales and the fact that mobile dislocations interact with SSD and GND in a similar way, plastic deformation in materials containing strain gradients can be described in a satisfactory manner [51].

The third area is the surface grain model which is a typical shape sum effect: the measured yield strength is represented as the sum of a portion from the core grains and from the surface grains. The latter contribute less to strain hardening, since they are only sparsely integrated in the crystal structure. Typical sample sizes are between 10 and 100 μm , respectively, less than about 20 times the typical grain size. The yield strength decreases with a decreasing number of grains per cross-section, because the portion of surface grains increases in return [22].

3.3.3 Yield Strength Models

Plasticity is due to dislocation movement on slip planes and influenced by different kinds of obstacle, which are either acting travel path limiting or can be overcome. The correlations between deformation mechanisms and yield strengths are well explained by Macherauch and Vöhringer [33]: In principal the yield strength σ_f can be divided into an athermal and a thermal contribution:

$$\sigma_f = \sigma_{\text{ath}}(\text{structure}) + \sigma_{\text{th}}(T, \dot{\varepsilon}, \text{structure}) \quad (3.4)$$

Both contributions are affected by the metallic microstructure, while the thermal part σ_{th} is additionally dependent on temperature T and strain rate $\dot{\epsilon}$.

The athermal contribution σ_{ath} is a result of four shares of $\Delta\sigma_{ath}^i$ and their corresponding mechanisms:

1. Dislocation hardening, which is based on the movement of glide dislocations within internal stress fields induced by other dislocations. The appropriate share to σ_{ath} is

$$\Delta\sigma_{ath}^1 = \sigma_{dis} = \alpha_{dis} \cdot G \cdot b \nu \cdot \sqrt{\rho_{dis}} \quad (3.5)$$

with a constant α_{dis} , the shear modulus G , the absolute value of the Burgers vector $b\nu$ and the dislocation density ρ_{dis} .

2. Grain boundary hardening, which is caused by glide dislocations accumulating at grain boundaries and twin boundaries. This contribution is strongly shaped by the grain size d_G and is known as the Hall–Petch relationship [15]:

$$\Delta\sigma_{ath}^2 = \sigma_{Gb} = GBR/\sqrt{d_G} \quad (3.6)$$

At this, GBR is a material dependent parameter with the significance of grain boundary resistance.

3. Solid solution hardening appears with respect to solved impurity atoms. It is based on elastic interaction between glide dislocations and foreign atoms, which are taken up in the crystal lattice in the direct neighborhood of the glide planes. At a foreign atom concentration c_{fa} (at.%) this leads to a portion

$$\Delta\sigma_{ath}^3 = \sigma_{ss} = \alpha_{ss} \cdot G \cdot c_{fa}^n \quad (3.7)$$

Thereby α_{ss} is a constant, G is the shear modulus and n an exponent, ranging between 0.5 and 1.

4. The contribution of all kinds of particle hardening is included in $\Delta\sigma_{ath}^4 = \sigma_p$. Particle hardening comprises interactions between glide dislocations, coherent and incoherent precipitations or dispersions acting as obstacles. Examples for particle hardening are the Orowan mechanism and the Kelly-Fine mechanism. Similar to grain boundary hardening, the presence of a grainy, hard second phase leads to an increase in yield strength.

If different hardening mechanisms are active at the same time, the yield strength can be estimated, using the additivity of σ_{ath} :

$$\sigma_{ath} = \sum_{i=1}^4 \sigma_{ath}^i = \sigma_{dis} + \sigma_{Gb} + \sigma_{ss} + \sigma_p \quad (3.8)$$

The contribution of material temperature and strain rate is included in the second term of Eq. (3.4), whereby the temperature dependence of σ_{th} is much greater than that of the strain rate. In contrast to body-centered cubic (bcc) metals, which show a pronounced temperature influence on yield strength, face-centered cubic (fcc)

metals do not do so. On the other hand, in fcc metals the hardening behavior is strongly temperature dependent as compared to bcc metals.

To quantify the effect of temperature T and strain rate $\dot{\epsilon}$ on the thermal contribution σ_{th} , the following relation is used:

$$\dot{\epsilon} = \frac{1}{m_{\text{Taylor}}} \cdot \rho_{\text{gd}} \cdot bv \cdot \frac{dL}{dt} \quad (3.9)$$

whereat the strain rate $\dot{\epsilon}$ is determined by the mean gliding path dL within the time interval dt , the absolute value of the Burgers vector bv and the density of glide dislocations ρ_{gd} . The grain accommodation factor m_{Taylor} provides the link between macroscopic and microscopic deformations of the polycrystal's grains (see also Sect. 3.3.4). The time interval dt comprises the running time of a glide dislocation between the short-range obstacles t_L on the one hand and a proportion of delay time t_D on the other hand, whereby it applies that $t_D \gg t_L$. If dL is now substituted by the mean distance between the short-range obstacles $\bar{L}$ it follows from Eq. (3.9) that

$$dt = t_L + t_D = \frac{\rho_{\text{gd}} \cdot bv \cdot \bar{L}}{m_{\text{Taylor}} \cdot \dot{\epsilon}} \quad (3.10)$$

The mean delay time t_D is determined by the probability of a local incidence of a sufficient fluctuation of the activation enthalpy ΔGFE which is known from statistical mechanics as

$$t_D = \frac{1}{f_{\text{Debye}}} \cdot \exp\left(-\frac{\Delta GFE}{K_B \cdot T}\right) \quad (3.11)$$

Thereby f_{Debye} is the Debye frequency, K_B the Boltzmann constant and T the absolute temperature. Using $t_D \gg t_L$ it can be concluded from Eqs. (3.10) and (3.11) that

$$\dot{\epsilon} = \dot{\epsilon}_0 \exp\left(-\frac{\Delta GFE}{K_B \cdot T}\right) \quad (3.12)$$

with a strain rate constant $\dot{\epsilon}_0 = \rho_{\text{gd}} \cdot bv \cdot \bar{L} \cdot f_{\text{Debye}} / m_{\text{Taylor}}$.

The vivid meaning of ΔGFE is reflected in Fig. 3.18. Force-distance curves are depicted of a short-range dislocation obstacle during material loading at different temperatures. In each case the force or, alternatively, the stress required for local dislocation movement is drawn schematically in dependence of the distance. At $T = 0$ K the force $F_{th,0}$ or the corresponding stress $\sigma_{th,0}$ have to be applied to overcome the obstacle $C_{th,0}$, since there are no thermal fluctuations at the absolute zero point. Micro metal forming normally takes place at ambient temperature, being in qualitative accordance with the temperatures T_1 or T_2 in Fig. 3.18. At these temperatures finite values of thermal energy ΔGFE_1 and ΔGFE_2 are available to overcome the obstacle, at which $\Delta GFE_2 > \Delta GFE_1$, because $T_2 > T_1$. The corresponding free enthalpies are marked as hatched areas in Fig. 3.18. It is obvious that the overcoming of the same obstacle resistances requires that the

larger the forces F_{th} or stresses σ_{th} , the lower the temperatures. Moreover one sees immediately that on reaching the temperature T_0 , the whole work necessary to overcome the obstacle is deposited thermally. Hence, the whole thermal material resistance $C_{th} = 0$ and consequently $F_{th} = 0$ and $\sigma_{th} = 0$. The required activation enthalpy is called ΔGFE_0 and is characteristic of the present type of obstacle.

It becomes obvious from Fig. 3.18 that ΔGFE is determined by F_{th} or σ_{th} . For many metals $\Delta GFE(\sigma_{th})$ can be approximated by a power law of the form

$$\Delta GFE = \Delta GFE_0 \left[1 - \left(\frac{\sigma_{th}}{\sigma_{th,0}} \right)^{1/m_{th}} \right]^{1/n_{th}} \quad (3.13)$$

Typical exponents for metals are e.g. $m_{th} = n_{th} = 1$ for pure aluminum and some hexagonal metals, $m_{th} = 1/2$ and $n_{th} = 1$ for titanium alloys, $m_{th} = 2$ and $n_{th} = 1$ for pure iron, $m_{th} = 4$ and $n_{th} = 1$ for carbon steels as well as $m_{th} = 2$ and $n_{th} = 2/3$ for copper alloys.

Substitution of Eq. (3.13) into Eq. (3.12) and solving the equation for the thermal flow stress contribution for temperatures $T < T_0$ leads to

$$\sigma_{th} = \sigma_{th,0} \left[1 - \left(\frac{T}{T_0} \right)^{n_{th}} \right]^{m_{th}} \quad (3.14)$$

with

$$T_0 = \frac{\Delta GFE_0}{K_B \cdot \ln \frac{\dot{\epsilon}_0}{\dot{\epsilon}}} \quad (3.15)$$

As a result, the influence of temperature and strain rate on the thermal flow stress is quantitatively determined and the summarization of Eqs. (3.4), (3.8) and (3.14) gives the flow stress as the sum of the thermal and athermal contributions:

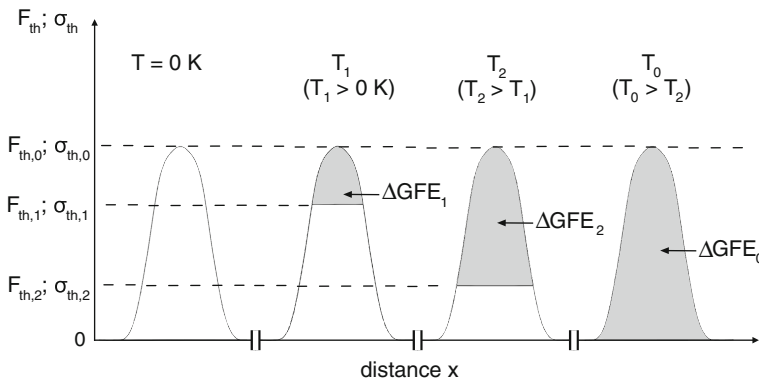
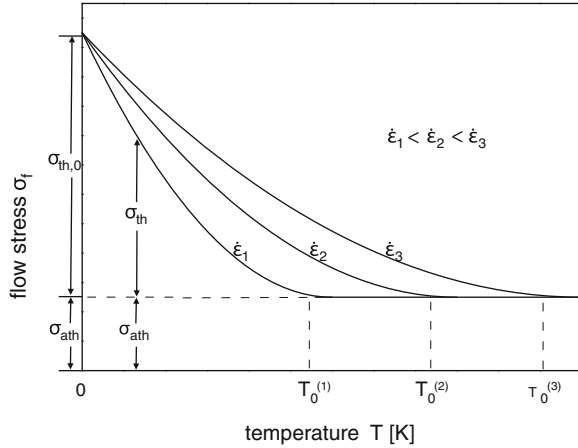


Fig. 3.18 Force-distance curves of a dislocation obstacle during material loading at different temperatures: micro cold forming usually takes place at ambient temperature being in qualitative accordance with the condition at T_1 or T_2 , according to Macherauch and Vöhringer [33], reproduced with permission from Wiley-VCH

Fig. 3.19 Influence of strain rate and temperature on the yield strength of polycrystals according to Macherauch and Vöhringer [33], reproduced with permission from Wiley-VCH



$$\sigma_f = \underbrace{\sigma_{\text{dis}} + \sigma_{\text{Gb}} + \sigma_{\text{ss}} + \sigma_p}_{\sigma_{\text{ath}}} + \underbrace{\sigma_{\text{th},0}(\text{structure}) \left[1 - \left(\frac{T}{T_0} \right)^{n_{\text{th}}} \right]^{m_{\text{th}}}}_{\sigma_{\text{th}}} \quad (3.16)$$

If the flow stress is calculated for different strain rates depending on temperature, schematically the curved shapes depicted in Fig. 3.19 result. The flow stresses continuously decrease with increasing temperature and are discharging earlier into σ_{th} the lower the strain rate.

Micro metal forming processes are usually carried out at low homologous temperatures $T^* = T/T_m$ with T_m = melting temperature, where deformation can be described on the basis of stress and strain. Every point on the strain hardening curve therefore equates to the flow stress σ_f of the instantaneous deformation state. Obviously time-dependent processes play no significant role at low temperatures. A fact which expresses itself in the dependence of the yield strength $\sigma_{\text{p}10.2}$ on the strain rate $\dot{\varepsilon}$ which can be described by the strain rate sensitivity m_f :

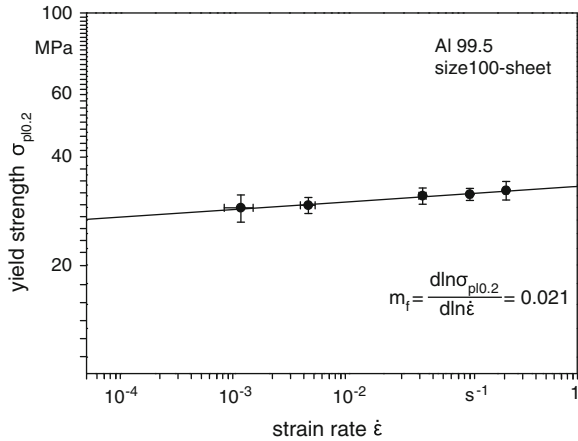
$$m_f = \frac{d \ln \sigma_{\text{p}10.2}}{d \ln \dot{\varepsilon}} \quad (3.17)$$

For example, typical values for the strain rate sensitivity of fcc metals are approximately $m_f \cong 1/100$ as can be seen in the example for size 100 Al-sheets depicted in Fig. 3.20. Therefore σ_f is nearly independent of the strain rate. The definition of strain rate sensitivity m_f in Eq. (3.17) has been chosen because the dependency $\sigma_{\text{p}10.2}(\dot{\varepsilon})$ can usually be described approximately by a power law

$$\sigma_{\text{p}10.2} = K \cdot \dot{\varepsilon}^m \quad (3.18)$$

where $K = K(\varepsilon)$.

Fig. 3.20 Influence of strain rate $\dot{\epsilon}$ on yield strength $\sigma_{p10.2}$ described by the strain rate sensitivity m_f on the example of size 100 Al-sheets



3.3.4 Size Effects on Yield Strength

Material structures are basically subdivided into polycrystalline and single crystalline materials. While the plastic behavior of a single crystal is significantly determined by its orientation to the tensile direction, its active slip systems and the critical shear stress. A polycrystal with many grains within the cross-section of a specimen shows a plastic behavior, which is also characterized by the averaged properties of all grains and their interaction among each other. Against this background, in the first half of the 20th century several efforts were taken in an attempt to deduce the laws of plasticity of polycrystalline materials from the well-known plastic behavior of single crystals. During this two size effects came to light, increasing the flow stress of a polycrystal: a “specimen size” and a “grain size” effect [2]. Both are significantly dependent on the number of grains per cross-section.

If specimens with an increasing number of grains are considered, coming from a single crystal, primarily the “specimen size” effect appears which increases the flow stress σ_f for two reasons: on the one hand the critical shear stress now has to be reached in all the grains contributing to plastic flow within the cross-section, and on the other hand strain continuity at the grain boundaries has to be ensured. In fields of less than about 20 grains per cross-section, it has been concluded that

$$2 \cdot \tau_{sc} \leq \sigma_f \leq m_{or} \cdot \tau_{sc} \quad (3.19)$$

whereby the flow stress may vary between two times the single crystal critical shear stress τ_{sc} , for a structure containing at least one grain within the specimen’s cross-section, to m_{or} times the single crystal strength, at which the lower and upper levels of m_{or} are determined by Sachs’ or Taylor’s polycrystalline models, respectively [43, 48].

Sachs introduces an average crystal orientation factor $m_{\text{Sachs}} = 2.238$, which has been calculated under the assumptions that the axes of grains are orientated randomly, no interaction between adjacent grains takes place and that each crystal grain is subjected to sufficient uniaxial stress in the direction of the macroscopic stress to make it yield. His model represents a lower estimate for m_{or} , because frictional effects between the grains have not been considered.

Taylor's approach of calculating an upper limit for m_{or} is based on the principle of virtual work. From in total 12 slip systems of a fcc crystal system, five are always independent, leading to 384 possible combinations of five independent slip systems, each of them approving arbitrary deformation. Based on the simplified hypothesis that those slip systems are always chosen, whose overall shear

$$\tau_{\text{N}} = \sum_{i=1}^5 |\gamma_i| \quad (3.20)$$

is smallest, Taylor determines the work-hardening behavior of polycrystals. Assuming a random distribution of grain orientations, he calculates an upper bound $m_{\text{Taylor}} = 3.06$ for the grain accommodation factor m_{or} .

The “grain size” effect occurs if a sufficient number of grain boundaries is present in the polycrystal, roughly counting more than about 20 grains in the cross-section. These grain boundaries act as obstacles to the initiation of bulk yielding, and extra internal stress concentrations are required to induce a general yielding of the specimen. The “grain size” effect has a direct dependence on the grain size d_{G} and increases the yield strength, superimposing the “specimen size” contribution leading to Eq. (3.21), which is known as the Hall–Petch relation, with the “friction stress” σ_{i} and the “grain boundary resistance” GBR (see also Sect. 3.3.3):

$$\sigma_{\text{f}} = m_{\text{or}} \cdot \tau_{\text{sc}} + \frac{GBR}{\sqrt{d_{\text{G}}}} = \sigma_{\text{i}} + \frac{GBR}{\sqrt{d_{\text{G}}}} \quad (3.21)$$

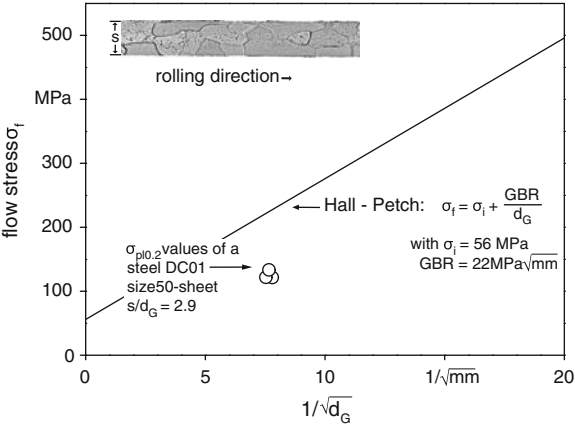
The Hall–Petch equation well describes the relation between grain size and flow stress for conventional specimens, but it does not hold true for small ones, such as the size50-sheet of steel DC01 shown in Fig. 3.21: The measured flow stress $\sigma_{\text{p}0.2}$ of these sheets is about 45 % lower than would be expected on the basis of the Hall–Petch relation for normal sized specimens [19]. At the same time the investigated sheets have a median grain size of 17 μm leading to a thickness to grain size ratio s/d_{G} of 2.9. As a consequence, about 75 % of the grains are in direct contact with the surface of the sheet.

An extended Hall–Petch relationship accommodating this derivation, by introducing two parameters m_{or}^{α} and β , is given by Kim et al. [26]:

$$\sigma_{\text{f}} = m_{\text{or}}^{\alpha} \cdot \tau_{\text{sc}} + \frac{GBR}{\sqrt{d_{\text{G}}}} \cdot \beta \quad (3.22)$$

At this m_{or} is the crystal orientation factor within the meaning of Eq. (3.19) and α and β are size-dependent parameters. The one limiting case, having a single

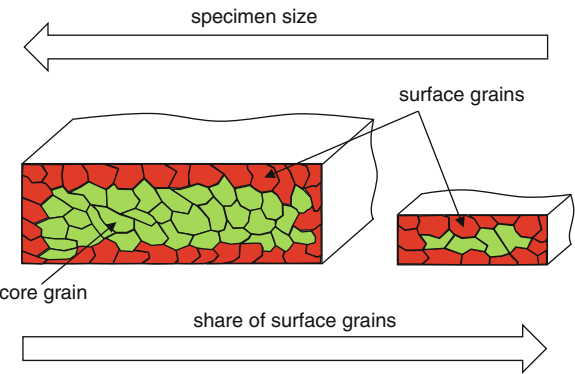
Fig. 3.21 Hall-Petch relation for conventional sized specimens compared with experimental data of a size50-sheet of steel DC01



crystal, leads to the lower bound conditions $m_{or}^{\alpha} \geq 2$ due to Eq. (3.19) and $\beta = 0$, since there are no inner grain boundaries responsible for a stress increase. The other limiting case, having a polycrystal, leads to the upper boundary condition $\alpha = \beta = 1$ transforming Eq. (3.22) into Eq. (3.21). As an example, the orientation factor m_{or} for the fcc structure of aluminum and copper has been determined experimentally to approximately $m_{or} = 2.6$, about halfway between the values reported by the Sachs and Taylor models [26]. The validity of the extended Hall-Petch relation could be proven experimentally for a large range, from fully polycrystal specimens down to parts that are only contain several grains.

The limited validity of the conventional Hall-Petch relationship [Eq. (3.21)] in the sub-millimeter range is due to the shape sum effect, a size effect where the observed value is depicted as a sum of core-dominated and surface-dominated subvalues, while the relative amounts of the two change due to scaling [50]. An example of a shape sum effect is the surface grain model. It is shown that neither the absolute grain size nor the absolute sample size are the determining quantities, but rather the ratio of grain size to cross-section, or in other words the volume ratio of surface grains to core grains. Figure 3.22 depicts the increase in the share of

Fig. 3.22 Correlation between specimen size and share of surface grains according to Kals et al. [22]



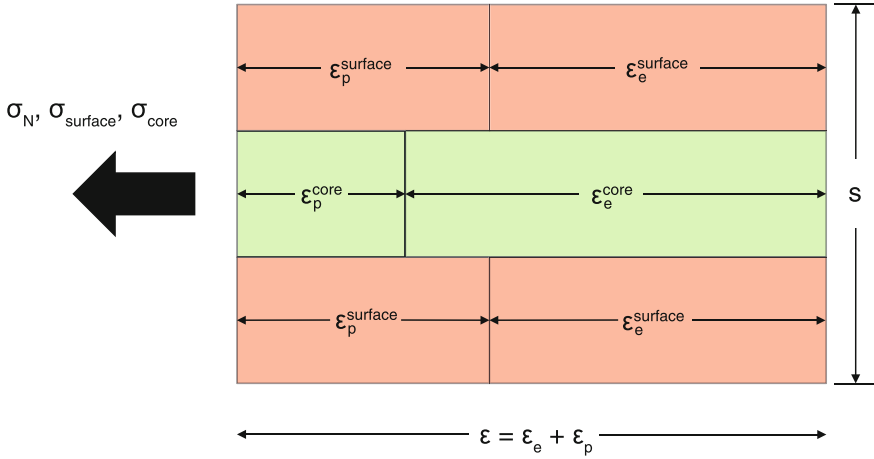


Fig. 3.23 Schematic drawing of stress and strain behavior within the surface zones and the core zone

surface grains when the specimen size decreases while the grain size stays the same.

The grains lying at the surface make a smaller contribution to strain hardening than do the core grains, since the latter are completely integrated in the crystal structure. From the view of hardening mechanisms the dislocation movement in surface grains is less obstructed than it is in core grains, because dislocations moving through the grains during deformation pile up at grain boundaries, but not at the free surface, where no extra internal stress concentrations are required to overcome grain boundaries [12]. Against this background, it is obvious to differentiate model-like between the surface zone and a core zone, attributing a different behavior with various elastic and plastic shares to each of them (Fig. 3.23) and, as a result, assigning a flow stress $\sigma_f^{\text{surface}}$ adding the lower part of the surface and a flow stress σ_f^{core} contributing to the higher ones for the core of the sample. Within this model the measured flow stress σ_f can be represented as a linear combination of the form

$$\sigma_f = \theta \cdot \sigma_f^{\text{surface}} + (1 - \theta) \cdot \sigma_f^{\text{core}} \quad (3.23)$$

with the volume fractions θ and $1 - \theta$ of the surface or the core, respectively. If in particular only surface grains are present ($\theta = 1$) and the specimen size will be reduced further, the flow stress no longer decreases [22]. Associated therewith, the orientations of single grains become more important and the scatter of flow stress of different samples increases.

Thus the observed deviations from yield strength can be well explained due to the surface grain model as a shape sum effect which becomes increasing significant the larger the volume fraction of the surface zone becomes.

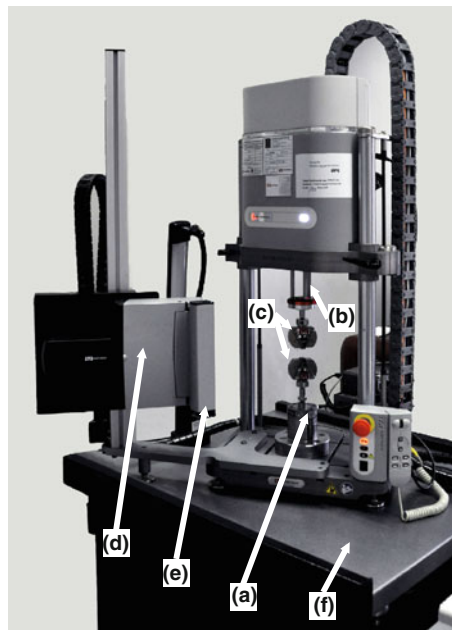
3.3.5 Requirements for Testing Equipment

The mechanical characterization, tensile or compression tests, of e.g. specimens with sub-millimeter dimensions necessitates testing equipment capable of meeting specific requirements: The testing system has to allow low forces within the range of a few Newton and adequate piston strokes to be set and controlled with sufficient accuracy. Figure 3.24 shows such a testing system which covers the low force range from 0 to $\pm 1,000$ N and also permits static and cyclic tests.

Especially within the low force range, environmental issues such as vibrations or footsteps can thwart the measurement. Therefore it is advisable to decouple the testing system mechanically from its environment using an adequate vibration isolation system, for example a floating laboratory bench.

Moreover, for specimens with dimensions in the sub-millimeter range contact strain sensors are unsuitable because their interaction with the sample is not negligible. Instead of that the strain has to be detected with non-contact methods such as capacitive or inductive displacement sensors or by means of optical strain measurement methods, like grey-scale correlation, and laser- or video extensometers [18]. Such a video extensometer is exemplarily shown in Fig. 3.25: Two dots (c) are applied as markers within the gauge length on the front side of a tensile test specimen which is lit with polarized, monochromatic light by the illumination unit (a). The video camera (b) detects and follows the balance points of the markers.

Fig. 3.24 Electrodynamic testing system. **a** load cell, **b** piston, **c** clamping device, **d** video extensometer, **e** illumination unit, **f** floating bench



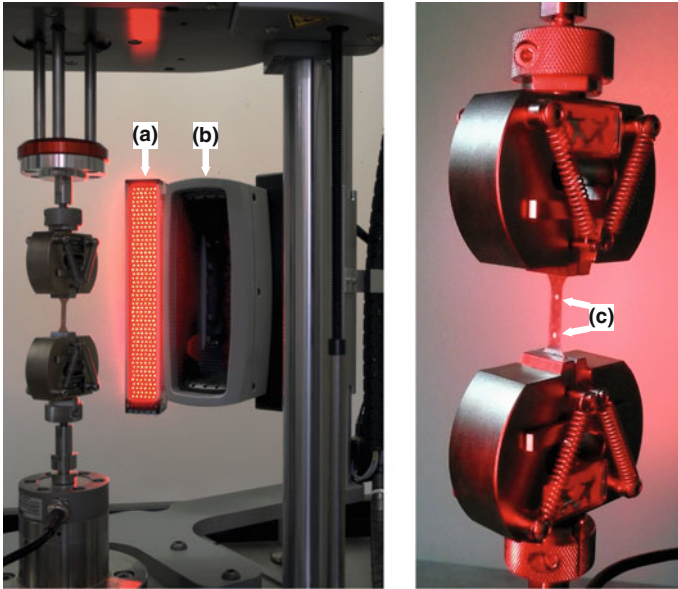


Fig. 3.25 Non-contact strain measurement with a video extensometer: The video camera (b) detects and follows the two markers (c) applied to the front side of the specimen

3.3.6 Flow Stress Atlas

The following is an exemplary collection of flow stress and stress–strain data from different investigations which give an overview and orientation to how specimen and grain size influence the flow stresses of some relevant materials Fig. 3.26 +

Fig. 3.26 Aluminum (99.999 %, rolled and heat treated). Stress–strain curves of Al foils with varying thickness according to Klein et al. [27]. Reprinted from Klein et al. [27] with permission from Elsevier

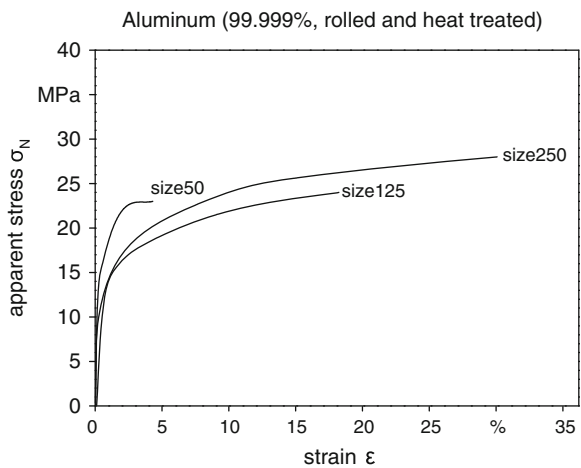


Fig. 3.27 Stress–strain curves of Al (99.5 %) foils with varying thickness according to own investigations and Zhang et al. [56]. Reprinted from Zhang et al. [56] with permission from Elsevier

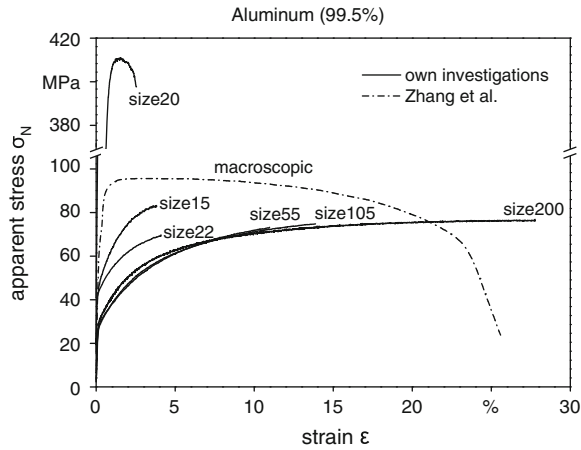
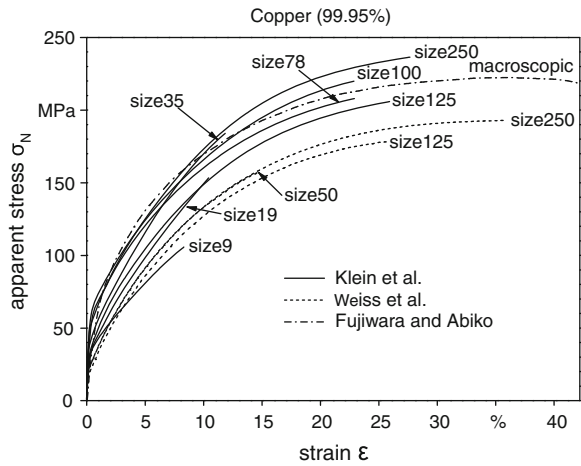


Fig. 3.28 Stress–strain curves of Cu (99.95 %) foils with varying thickness and of a macroscopic specimen for comparison according to Fujiwara and Abiko [10], Klein et al. [27], Weiss et al. [53]. Reprinted from Klein et al. [27] with permission from Elsevier. Reprinted from Weiss et al. [53] with permission from Elsevier. Reprinted from Fujiwara and Abiko [10] with permission from Journal de Physique (<http://www.journaldephysique.org>)



Tables 3.2; Fig. 3.27 + Table 3.3; Fig. 3.28 + Table 3.4; Fig. 3.29 + Table 3.5; Fig. 3.30 + Table 3.6; Fig. 3.31 + Table 3.7; Fig. 3.32 + Table 3.8; Fig. 3.33 (without Table).

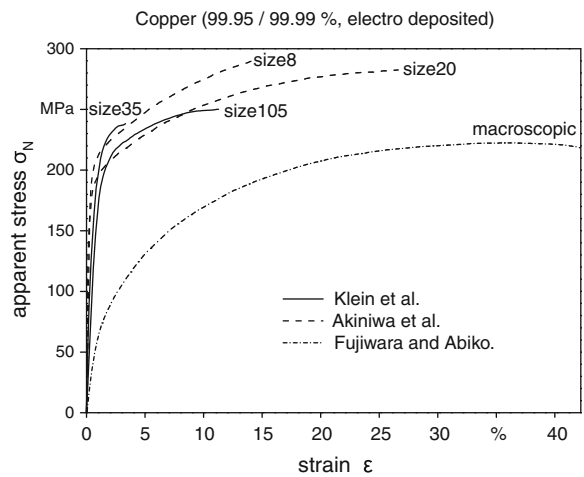


Fig. 3.29 Stress–strain curves of electrodeposited Cu (Klein et al. [27] used copper with 99.95 % purity, Akiniwa et al. [1] utilized a sample with 99.99 % purity), foils with varying thickness and of a macroscopic specimen for comparison according to Akiniwa et al. [1], Klein et al. [27], Fujiwara and Abiko [12]. Reprinted from Klein et al. [27] with permission from Elsevier, Reprinted from Akiniwa et al. [1] with permission from Trans Tech Publications and Y. Akiniwa, Reprinted from Fujiwara and Abiko [10] with permission from Journal de Physique (<http://www.journaldephysique.org>)

Fig. 3.30 Stress–strain curves of low carbon steel DC01: size50 specimens in different annealing conditions according to Köhler et al. [29] (Reprinted from Köhler et al. [29] with permission from Elsevier) and a macroscopic specimen for comparison according to Boller and Seeger [3]. Reprinted from Boller and Seeger [3] with permission from Elsevier and T. Seeger

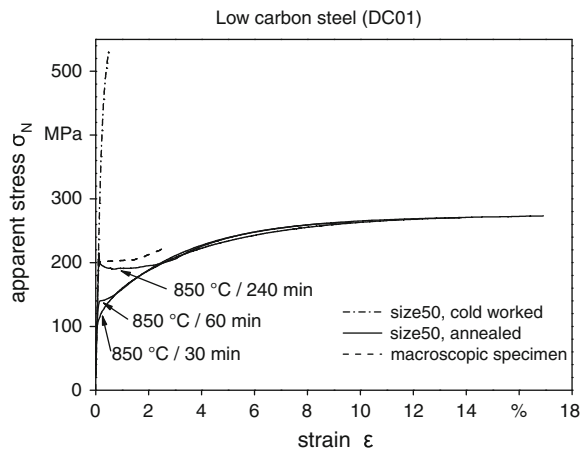


Fig. 3.31 Stress–deformation curves determined by the upsetting of geometrically similar cylindrical specimens of CuZn15 according to Meßner and Engel [37]. Reprinted from Meßner and Engel [37] with permission from U. Engel

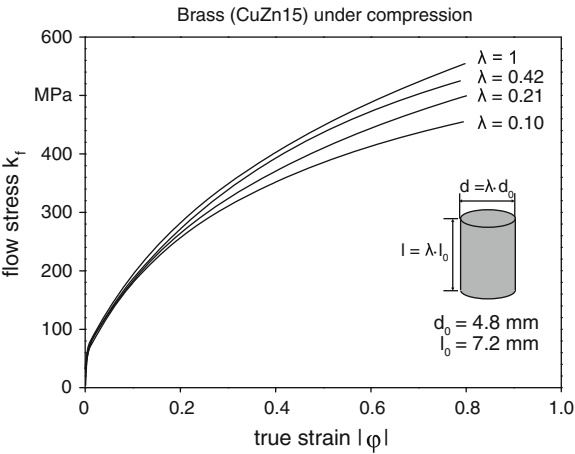


Fig. 3.32 Stress–strain curves of steel 1.4301 foils with varying thickness (own investigations)

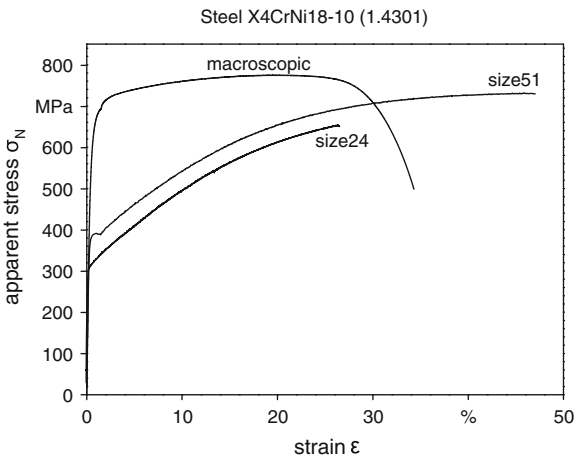


Fig. 3.33 Influence of the specimen size on the flow stress k_f for the steel 1.4301 determined by tensile tests, reprinted from Meßner and Engel [37] with permission from U. Engel

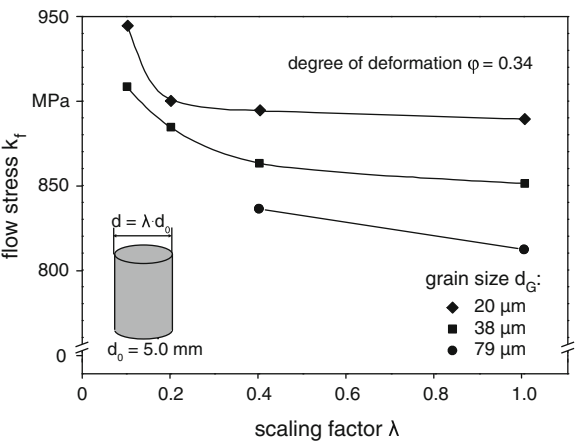


Table 3.2 Microstructural parameters and material properties

Condition	Young's modulus E (GPa)	Thickness s (μm)	Grain size d_G^a (μm)	s/d_G	σ_m (MPa)	ε_{fr} (%)	Source
Rolled and heat treated (550 °C/2 h)	69.1	50	140	0.36	23	4.2	[27]
	70.5	125	213	0.59	27	17.6	
	69.4	250	218	1.17	28	28.5	

^a The grain sizes have been determined by the linear intercept method

Table 3.3 Microstructural parameters and material properties

Condition	Thickness s (μm)	Grain size d_G (μm)	s/d_G	HV 0.01	$\sigma_{pl0.2}$ (MPa)	σ_m (MPa)	ε_{fr} (%)	Source
Rolled	15	–	–	46	50	82	3.2	Own investigations
	22	25	0.8	35	44	66	3.8	
	55	25	2.0	48	29	72	9.8	
	105	25	4.0	68	29	73	13.8	
	200	–	–	52	31	77	29.0	
PVD sheet	20	2 ^a	10.0	189	385	401	0.8	
		Diameter d (μm)	d/d_G					
Annealed, 360 °C 1/2 h	9,900	62	159.7	–	84	99	3.0	[56]

^a Typical PVD sheets are composed of columnar crystals perpendicular to the sheet surface. Within this context the diameter of the columnar crystals has been assumed as d_G

Table 3.4 Microstructural parameters and material properties of rolled copper

Condition	Young's modulus E (GPa)	Thickness s (μm)	Grain size d_G (μm)	s/d_G	σ_m (MPa)	ε_{fr} (%)	Source
Rolled and annealed (600 °C/2 h)	112	9	25	0.4	92	7.5	[27]
	82	19	15	1.3	124	10.5	
	132	35	20	1.8	157	12	
	106	78	30	2.6	180	23	
	112	100	35	2.9	192	22.5	
	99	125	40	3.1	178	26	
	104	250	45	5.56	204	27	
	90	50	45	1.1	161	15	[53]
	95	125	40	3.1	184	27	
	105	250	45	5.6	199	36	
		Diameter d (μm)		d/d_G			
Cold swaged and annealed (300 °C/½ h)	130	3,000	30	100	222	33	[10]

Table 3.5 Microstructural parameters and material properties of electrodeposited copper

Condition	Young's modulus E (GPa)	Thickness s (μm)	Grain size d_G (μm)	s/d_G	σ_m (MPa)	ε_{fr} (%)	Source
Electro-deposited	92	35	2–5	7–17.5	238	26	[27]
	102	105	7–10	10.5–15	255	8	
	–	8	3	2.67	289	14	[1]
	–	20	10	2	282	26.5	
		Diameter d (μm)		d/d_G			
Cold swaged and annealed (300 °C/½ h)	130	3,000	30	100	222	33	[10]

Table 3.6 Microstructural parameters and material properties of low carbon steel DC01

Condition	Thickness s (μm)	Grain size d_G (μm)	s/d_G	HV 0.01	$\sigma_{p10.2}$ (MPa)	σ_m (MPa)	ε_{fr} (%)	Source
Cold worked	50	7.6	6.6	251	504	509	1	[29]
Annealed 850 °C	30 min	50	2.8	146	124	264	14	
	60 min	50	3.2	143	138	264	12	
	240 min	50	2.9	154	177	246	8	
	Diameter d (μm)		d/l	d_G				
Hot rolled	6,000	–	–	–	207	359	–	[3]

Table 3.7 Microstructural parameters

Condition	Scaling factor λ	Initial length l (μm)	Initial diameter d (μm)	Grain size d_G (μm)	d/d_G	Source
Annealed	1	7,200	4,800	80	60	[37]
	0.42	3,000	2,016	80	25.2	
	0.21	1,500	1,008	80	12.6	
	0.10	750	480	80	6	

Explanatory note The upsetting experiments were performed using PTFE foils to minimize the interface friction between the lapped dies and the end surfaces of the specimens

Table 3.8 Microstructural parameters and material properties

Condition	Thickness s (μm)	Grain size d_G^a (μm)	s/d_G	HV 0.01	$\sigma_{\text{pl}0.2}$ (MPa)	σ_{m} (MPa)	ε_{fr} (%)	Source
Rolled	24	14.4	1.7	332	309	643	27	Own investigations
	51	6.4	8.0	336	380	736	45	
	Diameter d (μm)		d/d_G					
Hot rolled	10,000	63.3	160.0	468	576	775	34	Own investigations

^a The grain sizes have been determined by the area enumeration method

Fig. 3.34 Schematic $S-N$ curves with endurance and fatigue limit

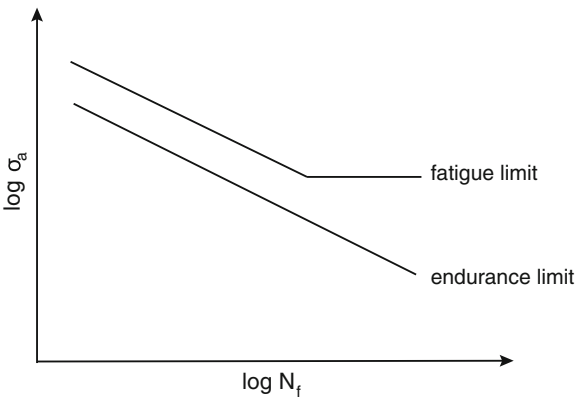


Fig. 3.35 Structure of a cardiovascular stent (Medtronic S7) after deployment, reprinted from Conolly et al. [6] with permission from Wiley–VCH

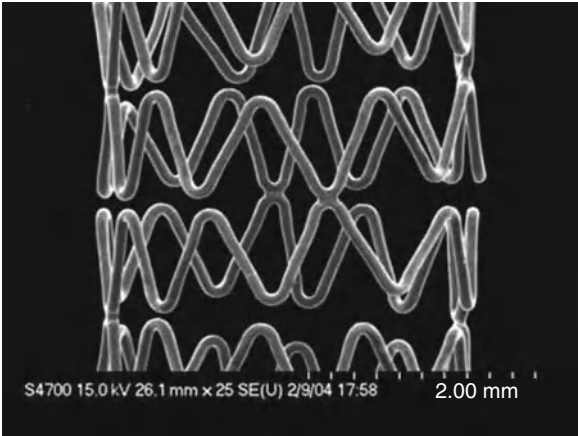
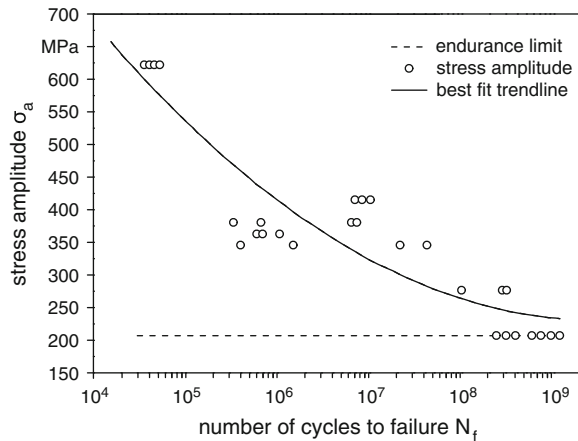


Fig. 3.36 S - N diagram for 125 μm diameter wire specimens of L-605 cobalt-chromium alloy, tested in rotary bending at $R_\sigma = -1$ in 37 $^\circ\text{C}$ 0.9 % saline solution at 60 Hz. Every point with a coordinate $N < 4 \times 10^8$ cycles represents a failed specimen. The other four points represent unfailed specimens, reprinted from Marrey et al. [34] with permission from Elsevier



3.4 Endurance Limit

Hubert Bomas and Bernd Köhler

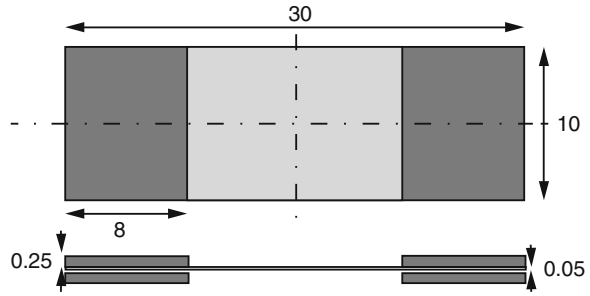
3.4.1 Definition

Metals show a specific reaction to repeated mechanical loading, this phenomenon is known as fatigue. Laboratory fatigue tests are most commonly conducted with constant load amplitudes which, on exceeding a certain value, may after a number of load cycles induce crack initiation, crack propagation and final fracture. The number of cycles until fracture is usually termed N_f . The fatigue regime depends on the lifetime divided roughly into three areas:

1. low-cycle fatigue (LCF), $N_f \leq 10^4$
2. high-cycle fatigue (HCF), $10^4 < N_f < 10^7$, and
3. very high-cycle fatigue (VHCF), $10^7 \leq N_f$

In the LCF regime the load in laboratory tests is usually applied as a strain amplitude, whereas in the other regimes it is applied as a stress amplitude. This chapter concentrates on the HCF regime and the loads that are applied as stresses. The results of tests with constant stress amplitudes are often presented as S - N or Woehler curves. Figure 3.34 shows two S - N curves ending in a fatigue limit and in an endurance limit, respectively. The fatigue limit is combined with distinct material behavior and thought to be the stress amplitude that is the border between the damaging and the non-damaging stress amplitudes at an infinite number of cycles [38]. The endurance limit describes another material behavior and is thought to be the stress amplitude that is the border between the damaging and the non-damaging stress amplitudes at a fixed large number of cycles. It must be mentioned that many authors use these definitions the other way round [40]. Both

Fig. 3.37 Fatigue specimen for axial loads made from a thin sheet with affixed heads



values are in a mathematical sense random variables, meaning that the curves in Fig. 3.34 have to be interpreted as locations with a failure probability of 50 %.

Endurance limit and fatigue limit are not always separated in the literature. In many cases discrimination is not possible; Fig. 3.35 shows the structure of a cardiovascular stent made from wires by a process which includes the cold forming processes of crimping and deployment [6]. Figure 3.36 shows an S – N curve of cobalt-chromium wires which are used for stents tested in saline solution. Every point represents one specimen. An enormous scatter of lifetimes is visible. The authors assume these wires have a fatigue limit. Statistical analysis under this assumption proves it to be 207 MPa [34].

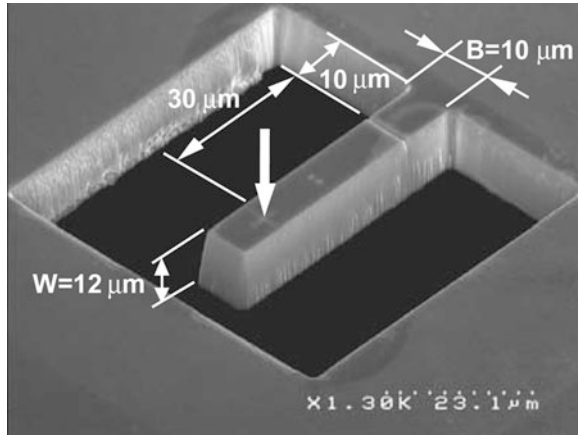
3.4.2 Measuring Techniques

Fatigue tests for the determination of the endurance limit are constant-amplitude tests which are also used for lifetime testing. The result of a constant-amplitude test is the lifetime N_f of the tested specimen, or the information that the specimen did not fail before reaching the limiting number of cycles, but gives no value for the endurance limit. A repetition of the test with other specimens at the same stress amplitude σ_a gives the ratio between the number of failed specimens and the number of tested specimens. This ratio serves as estimator for the failure probability $Pr_f(\sigma_a)$, which is identical to the cumulative distribution function CDF_e of the endurance limit:

$$CDF_e(\sigma_a) = Pr(\text{endurance limit} \leq \sigma_a) = Pr_f(\sigma_a) \quad (3.24)$$

Though constant-amplitude tests are a non-economic procedure to obtain the endurance limit, these tests are state of the art due to the extreme sensitivity of metal fatigue to load history. The distribution function of the endurance limit is often described by a three- or two-parametric Weibull distribution or by a log-normal distribution. The following equation gives an example for a two-parametric Weibull distribution with scale parameter σ_0 and shape parameter m_{Weibull} :

Fig. 3.38 Notched bending-fatigue specimen machined from Ni-P by focused ion beam technique, reprinted from Takashima and Higo [47] with permission from Wiley-VCH



$$CDF_e(\sigma_a) = 1 - \exp\left(-\left(\frac{\sigma_a}{\sigma_0}\right)^{m_{\text{Weibull}}}\right) \quad (3.25)$$

The scale parameter characterizes the 63 % quantile and depends on the material, the tested geometry, the type of loading, and the environment. The shape parameter characterizes the scatter of the endurance limit. A typical value for metals is $m_{\text{Weibull}} = 20$.

In order to determine the endurance limit, the applied testing machine should reach the necessary number of cycles in a reasonable time, meaning the load frequency should be about 20 s^{-1} or higher. Since the fatigue of metals can be a very selective process, which not only concerns the pure material but also the surface,

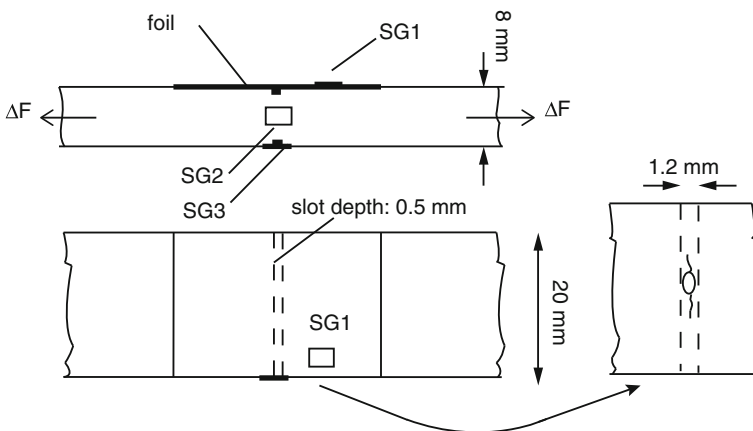


Fig. 3.39 Fatigue-testing device for a thin foil glued to a bar with a slot, SG: strain gauge, according to Klein et al. [27] with permission from Elsevier

size and preparation of the specimens, some care has to be taken with regard to specimen design and preparation. The mounting of a specimen on the testing machine usually induces stress concentrations in the specimen in the neighborhood of the specimen holder and can be reduced by the specimen geometry. Specimens that are to be axially loaded usually have heads to reduce the stress concentration (Fig. 3.37). This is not necessary if the stress concentration due to clamping falls in a region of low load stress or if the specimen has a notch (Fig. 3.38).

Another technique for handling thin foils is to glue them to a holder. Figure 3.39 shows such a device that has been applied to aluminum and copper foils. The foil is glued on a bar with a slot. The latter is situated underneath the foil and induces a strain increase in the part of the foil that bridges the slot. The right part of the figure indicates that crack initiation will be in this highly strained region.

3.4.3 Influences on the Endurance Limit

The influences on the endurance limit can be divided into four classes. These are influences from

1. the material,
2. the geometry,
3. the type of loading, and
4. the environment.

3.4.3.1 Material

Concerning the material, it is usually said that a high endurance limit requires a hard and tough material. Unfortunately, hardness measurement in micro-sized specimens is a task which requires some effort which is why most fatigue results of micro parts in the literature do not include the hardness of the materials. The

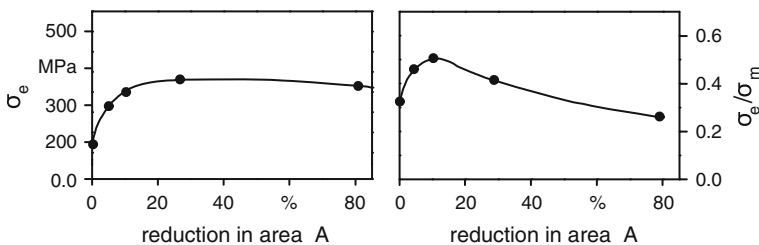


Fig. 3.40 *Left:* endurance limit at 30 million cycles of 2 mm wires of austenitic steel under rotating bending depending on the reduction in area due to hard-drawing, *right:* fatigue ratio σ_e/σ_m depending on the reduction in area. The surfaces of the wires are machined in the case of 0 % reduction or as-drawn, reprinted from Coquillet et al. [7] with permission from Wiley-VCH

absence of this information makes it difficult to compare fatigue results from different authors that have been obtained with the same type of material.

Cyclic loading seems to seek the weakest point in the material, which is the basis of the weakest-link idea. Nevertheless, the fatigue behavior of a metal is often characterized by the fatigue ratio σ_e/σ_m which describes the ratio between the push–pull endurance limit and the tensile strength. In most cases this value ranges between 0.2 and 0.5 (Fig. 3.40, right side). Lower values are observed in hard and brittle metals, whereas the larger ones are observed in soft and ductile metals. In general, material influences on the endurance limit can be very different and specific so that an all-embracing description would go beyond the scope of this chapter.

One exception shall be made with respect to micro metal forming. Metal forming below the recrystallization temperature has a considerable influence on the material state; it increases its dislocation density and it stretches the grains. Additionally, it may influence the surface roughness and the residual stress state. As an example, the results of Coquillet et al. observed in austenitic stainless steel wires with a diameter of 2 mm are presented in Fig. 3.40 [7]. The wires have been produced by cold drawing, starting from different input diameters to a final diameter of 2 mm. Thus different degrees of deformation have been achieved and are quantified by the reduction in the wires' cross-section. It was found that tensile strength increases steadily with an increasing degree of deformation. The endurance limit seems to have a maximum at a 30 % reduction of area. Coquillet et al. argue that the endurance limit is not only influenced by the static strength but also by a simultaneous effect of increasing sensitivity against surface defects with increasing static strength, an argument that is consistent with the statements in the previous paragraph.

3.4.3.2 Geometry

The geometry aspect covers items such as size, notches and surface topography. Since size is very important in the context of micro metal forming there will be a specific chapter on this topic. Notches are described with a stress concentration factor K_t which is defined as the ratio between the maximum local stress $\sigma_{\max}$ in the notch root and the nominal stress σ_N :

$$K_t = \frac{\sigma_{\max}}{\sigma_N} \quad (3.26)$$

The fatigue notch factor describes the ratio between the nominal endurance limits of a smooth specimen and a notched specimen:

$$K_f = \frac{\sigma_{e0}}{\sigma_{N,en}} \quad (3.27)$$

Introducing the stress concentration factor K_t and the local endurance limit σ_{en} in the notch root in this equation leads to

$$\sigma_{en} = \frac{K_t}{K_f} \cdot \sigma_{e0} \quad (3.28)$$

It is always observed that the local endurance limit in the notch root expressed as a local stress is larger than the endurance limit of a smooth specimen: $\sigma_{en} > \sigma_{e0}$ and $K_t > K_f$. This is often explained as a size effect, since the highly stressed volume and surface in the notched specimen is smaller than those of the smooth specimen.

Soft materials usually fail due to surface crack initiation. If there are no distinct surface flaws, the surface roughness is important for the endurance limit. The most important parameters for describing surface roughness are

- the arithmetic average height parameter R_a , also known as the center line average,
- the 10-point height parameter R_z defined as the average summation of the five highest peaks and the five lowest valleys,
- the maximum peak to valley height R_y , and
- the maximum height of profile parameter R_t defined as the distance measured between the highest peak and the lowest valley.

Novovic et al. [41] performed a literature review on the effect of machined surface topography and integrity on fatigue life, and examined different roughness parameters. They concluded that R_t and R_z are better used in determining fatigue performance than R_a , because these parameters represent the worst defects in the surface [Nov05].

3.4.3.3 Type of Loading

The type of loading includes mean stresses, residual stresses, multiaxial stresses and stress gradients. The different stress types are considered by fatigue criteria from which equivalent stress amplitudes are derived. An equivalent stress amplitude describes a push-pull amplitude which has the same effect as the regarded complex cyclic load. Today, there is a lack of experiments on micro parts that investigate the influence of the stress state on the endurance limit. Consequently, no recommendations exist regarding which fatigue criteria can be applied successfully in the micro regime.

Stress gradients due to loading concentrate the local stress in a component or specimen in a small volume, as do notches. Typical load situations with stress gradients are bending or torsion. The effect on the endurance limit is the same as that of a notch and has already been described.

3.4.3.4 Environment

Environment summarizes the effects from surrounding media and temperatures on the endurance limit which may be very complex.

3.4.4 *Size Effects on the Endurance Limit*

The endurance limit of parts with conventional sizes decreases with an increasing size of the stressed volume or the stressed surface. This can be called a negative size effect. The reasons are classically divided into geometrical, statistical and technological size effects [28]. The geometrical size effect refers to the influence of the stress gradient on the endurance limit. The statistical size effect is based on the idea that the material contains defects and that in the case of high-cycle fatigue a crack would initiate at the most serious defect. With increasing size of the highly stressed volume the probability of including a critical defect will also increase. This is the basic idea of the weakest-link concept. The technological size effect summarizes different technological reasons that make a large specimen or component more susceptible to damage from cyclic loads than a smaller one. For wrought materials, one of these reasons, and in many cases the most important, is the degree of deformation during hot or cold processing, which usually increases with decreasing size. The influence of the degree of deformation is also present in cold-formed micro parts but is often concealed by other effects, as will be shown in the following.

In production engineering different categories of size effects have been established which seem to take more account of metal parts with dimensions that range in the region of sizes of microstructural material features, for example inclusions, dislocations or grains [50]. These are the density, shape, and micro structure size effects which have been characterized in detail in [Chap. 2](#).

3.4.4.1 Density Size Effect

The density or statistical size effect is of great importance for normal-sized parts, but seems to lose its significance in the case of micro parts. This is due to the greater importance of surface effects or processes that are active in grains which have contact with the surface. The following paragraphs give some examples for cold-formed and recrystallized materials.

3.4.4.2 Shape Size Effect

Figure 3.41 shows an example for the shape size effect; Judelewicz et al. produced recrystallized sheets of high-purity copper with thicknesses of 20 and 100 μm and

Fig. 3.41 S – N curves of recrystallized high-purity copper sheets under repeated tension with foil thicknesses of 20 and 100 μm , $R_\sigma = 0$, reprinted from Judelevic et al. [21] with permission from Elsevier

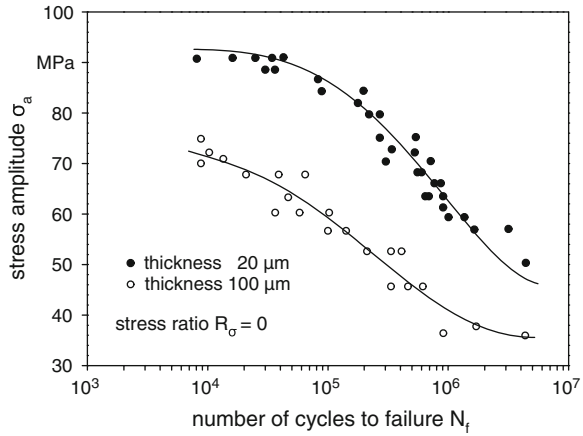
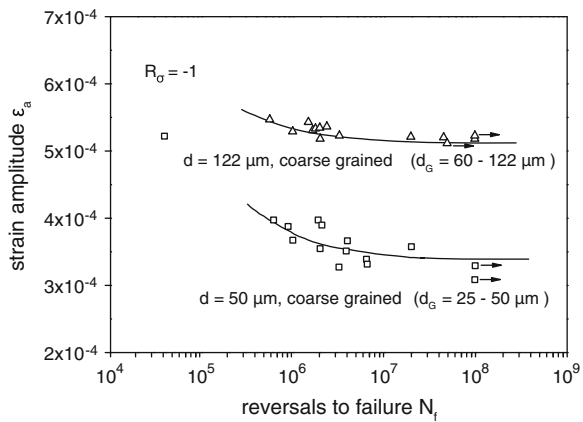


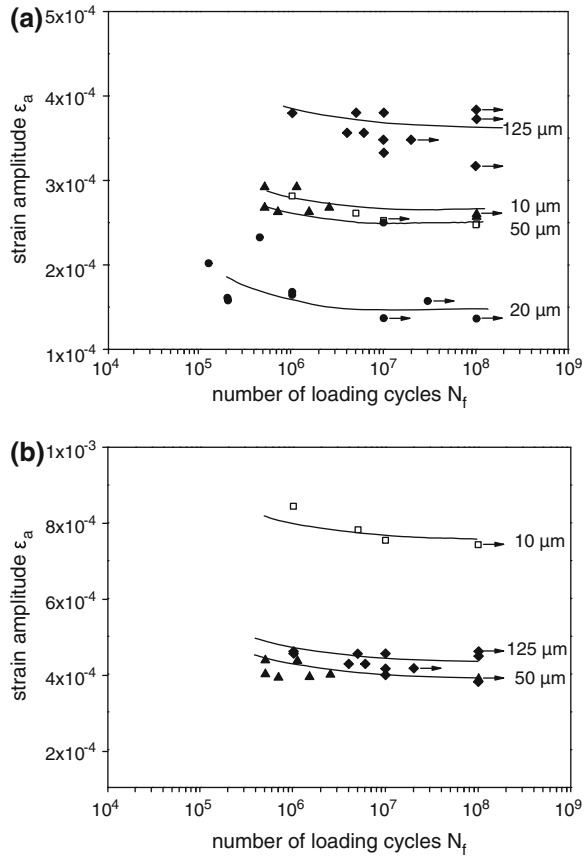
Fig. 3.42 S – N curves of annealed copper wires with different diameters d under tension-compression, d_G = grain size, reprinted from Weiss et al. [53] with permission from Elsevier



prepared fatigue specimens from them [21]. The grain size was in the order of 100 μm , meaning that almost all grains had contact with the surface and that in the case of the thin specimens a Tiffany grain structure was at hand. After a certain number of cycles the thicker specimens contained a large number of slip bands with, in some cases, well-developed extrusions which are responsible for fatigue crack initiation. The thinner specimens only contained slip bands in a small number of grains. It is concluded that the short migration path to the surface and the possibility of dislocation annihilation in the thinner specimens was responsible for this occurrence. Overall, this example shows the negative influence of size which is also known from larger parts, but the observed mechanism is only relevant in very thin specimens.

It is interesting to see that conditions, which at first glance seem to be very similar to the example before, lead to a positive influence of size; Fig. 3.42 shows the S – N curves of annealed copper wires with equal grain-size-to-thickness ratios

Fig. 3.43 Lifetime of copper wires depending on the strain amplitude applied and their diameter, **a** integral strain, **b** local strain, reprinted from Khatibi et al. [25] with permission from Elsevier



where the thicker wire exhibits a better endurance limit. The authors assume a reduced number of gliding systems in the thinner wire to be responsible for the reduced fatigue strength [53]. Compared with the example before, it seems to be of importance that the grains are smaller than the diameter of the wires, so that the effect of unimpeded movement of dislocations to the surface and the following annihilation is suppressed.

Another example in this sequence, which seems to have more similarity with the example given in Fig. 3.41, was produced by Khabiti et al., also on high-purity copper wires [25]. The wires had diameters from 20 to 125 μm and had been annealed in order to produce a bamboo grain structure that is similar to the Tiffany grain structure of sheets. The testing was carried out on a piezo-electric system with a stress ratio $R_\sigma = -1$ and a frequency of $20,000 \text{ s}^{-1}$.

Figure 3.43a shows the lifetime of the specimens depending on their thickness and the overall strain amplitude in the gauge length. Since there has been strain localization, Fig. 3.43b shows the results depending on the maximum local strain

Fig. 3.44 *S*–*N* diagrams of as-rolled high-purity copper sheets under repeated tension tested along and across the rolling direction, **a** sheet thickness = 100 μm, **b** sheet thickness = 20 μm, reprinted from Judelevic et al. [21] with permission from Elsevier

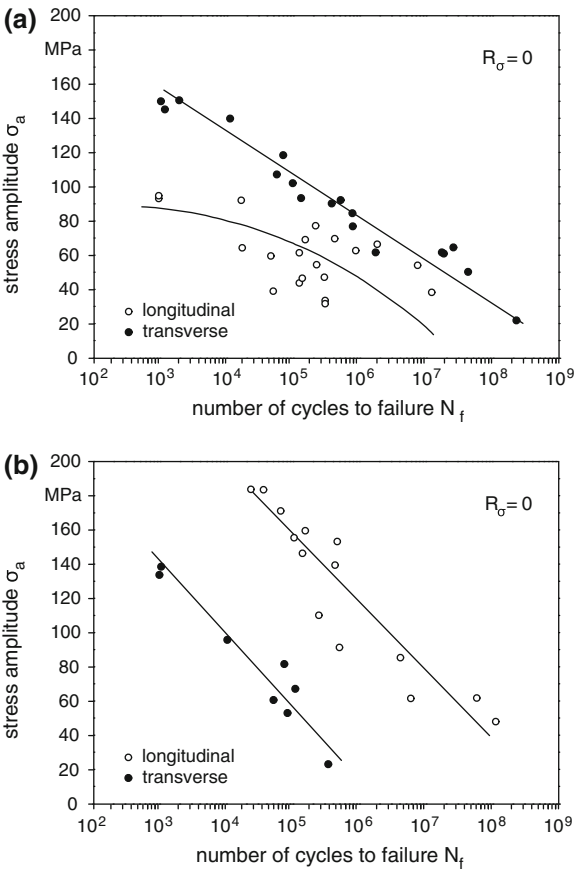
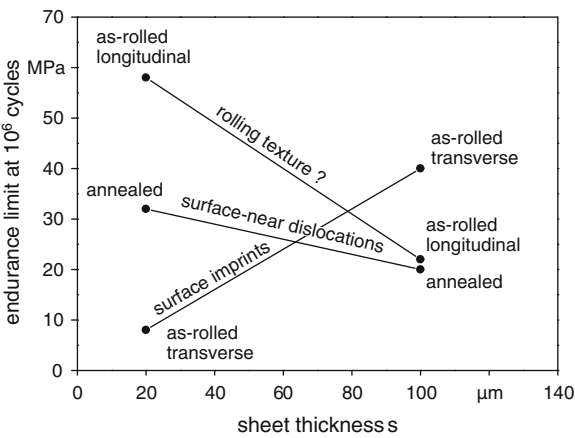


Fig. 3.45 Endurance limits of high-purity copper sheets according to the data of Judelewicz et al. [21] depending on sheet thickness under varying material conditions and testing directions, $R_\sigma = 0$



amplitude in the test length. Looking at the diagram with the local strain amplitude there seems to be a minimum endurance limit for a wire diameter of 50 μm . The authors discuss the dislocation structure in the wires and pay attention to a thin oxide layer at the surface of the wires which they believe impedes fatigue crack initiation.

3.4.4.3 Microstructure Size Effect

An example of the microstructure size effect is given by Hildebrand et al. They tested wires of the Ni–Co alloy MP35N with diameters between 127 and 229 μm in the LCF regime and found that fatigue strength increases with increasing diameter [17]. The reason for this effect has been determined by other authors as the presence of primary TiN particles of about 10 μm in size that do not change during the processing of the material [4]. In the case of the thin wires, these particles are often very near to the surface and their negative impact on fatigue behavior increases.

Another example has been published by Judelewicz et al. [21]. Figure 3.44 shows S – N diagrams of as-rolled sheets of high-purity copper. Figure 3.44a shows the results for a sheet thickness of 100 μm , and Fig. 3.44b shows them for a thickness of 20 μm . Both sheet types have rolling imprints of the same depth and direction. The result from the thicker sheet is explained by the rolling texture which produces a high number of glide systems with high Schmid factors if the stress is applied in the rolling direction. This leads to the relatively poor fatigue strength of the “longitudinal” specimens. In case of the thin specimens the “transverse” ones present poorer fatigue properties. This is due to the rolling imprints which are of the same size and direction as in the thicker specimens but cover a relatively larger area of the thin sheets’ cross-section.

The proceeding examples show that a size effect in the micro-regime may have different origins and directions. The experiments of Judelewicz et al. [21] cover an interesting range of conditions, and the measured endurance limits are worth being summarized in an extra figure (Fig. 3.45). It appears that further experiments and explanations are still necessary to determine the origins and understand the phenomena of size influence on fatigue behavior.

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Chapter 4

Bulk Metal Forming

Bernd Kuhfuss

Latin

A	Area (mm^2)
A_1/A_0	Cross section before and after the forming (mm^2)
d	Diameter (mm)
F	Force (N)
F_p	Punch force (N)
f_{st}	Stroke frequency (s^{-1})
g	Gap width (mm)
h	Height (mm)
h_p	Punch stroke (mm)
h_T	Effective tool stroke (mm)
I	Moment of inertia (mm^4)
k_f	Flow stress (MPa)
k_{fm}	Medium flow stress (MPa)
l_G	Grain size (μm)
NS	Neutral surface
r	Radius (mm)
u	Upset ratio
V	Formed volume (mm^3)
W_{id}	Ideal work (J)
W_p	Machine work (J)

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W_{pl}	Work of plunger (J)
Δx	Buckling (mm)

Greek

η_F	Efficiency of deformation
φ	Deformation degree
$\Delta\varphi_v$	Variance in the effective strain

4.1 Incremental Forming

Bernd Kuhfuss and Eric Mouri

4.1.1 Definition and Technologies

A common definition of incremental forming is given by Groche et al. [10, 11] in which an incremental forming process is characterized by regions of the workpiece experiencing more than one loading and unloading cycle due to the action of one set of tools within one production stage.

Incremental processes are divided into sheet metal and bulk metal forming. Well-established incremental bulk forming processes are open die forging, rotary swaging, flow forming and rolling processes (Fig. 4.1). In the following, only bulk metal forming with a focus on cold forming—namely rotary swaging—will be addressed.

Parts that are produced by rotary swaging are used as macro scale products in different applications. An important area of application are components for the automotive industry, like axes, steering spindles and gear shafts, in particular those made from tubular blanks with the purpose of weight saving.

Basic research on rotary swaging started in the 1960s, and has continued since then on macro scale components. In the last few years the focus of scientific interest was extended to micro components [15, 16]. Fields of application for swaged micro parts are seen in e.g. medical mechatronic systems (minimal invasive diagnostics and surgery) and the automotive industry (miniature pumps and valves).

In order to get a better understanding of the relevant effects during micro forming, the process is in general considered as a system consisting of the components process, tools, material and machine respectively equipment [6]. In particular, the process characteristics and material conditions are referred to in the following sections. Special features of micro rotary swaging machines and tool design are addressed in 8.1.

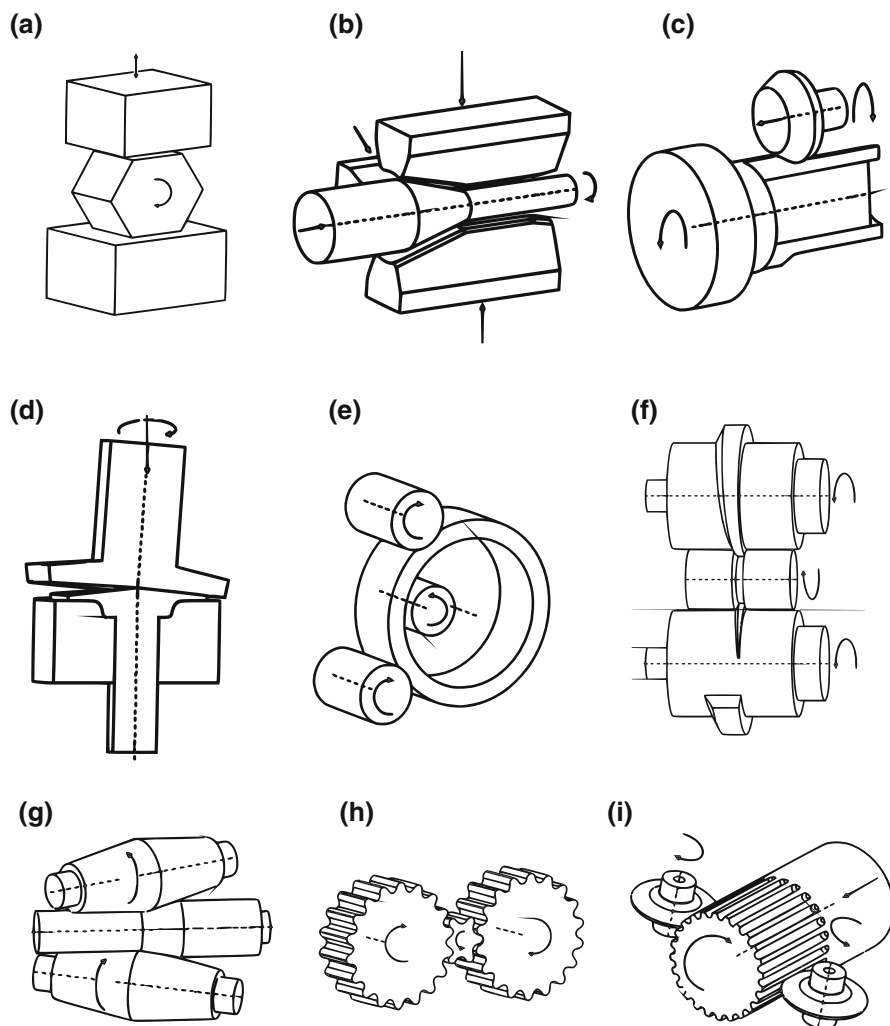


Fig. 4.1 Incremental forming processes according to [11]. **a** Open die forging, **b** rotary swaging, **c** flow forming, **d** orbital rolling, **e** ring rolling, **f** cross rolling, **g** skewed rolling, **h** profile rolling, **i** planetary rolling

4.1.2 Process and Tooling

According to DIN 8583, rotary swaging belongs to the processes of open die forging. The forming of the workpiece takes place in the swaging head in small steps by the radial oscillating movement of the tools (Fig. 4.2).

The radial movement of the forming tools (die segments) (1) is generated by the rotation of the driven shaft (2) and the sine-shaped cam on top of the base jaws (3).

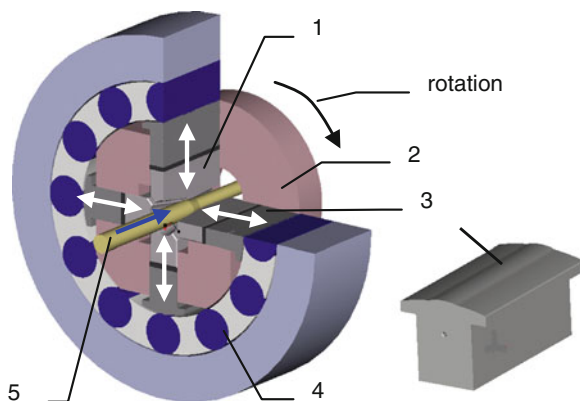


Fig. 4.2 Swaging head for the infeed process

With every pass of the cams along one of the cylinder rollers (4) the base jaws, and thereby the forming dies, are forced radially inwards producing an impact on the work (5). The rotation of the main shaft in relation to the workpiece leads to uniform forming over the circumference.

Rotary swaging can be sub-divided into two process variants with respect to the direction of feed. Figure 4.3 shows schematically the infeed swaging (left) and the plunge swaging (right) processes.

In infeed swaging the workpiece (1) is axially fed into the swaging head and reduction takes place over the whole feed length in the reducing zone of the dies (2). In plunge swaging the oscillating movement of the dies is superimposed by a radial

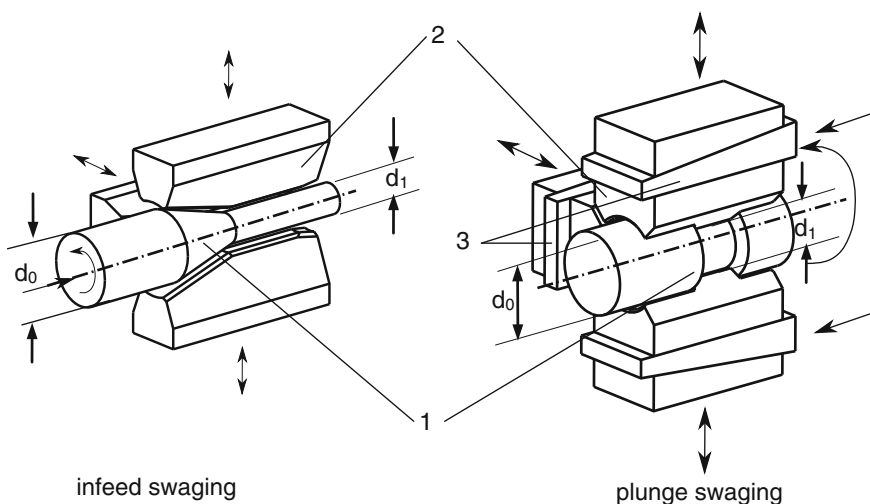


Fig. 4.3 Process variations of rotary swaging

feed rate due to the axial movement of additional wedges (3), so that a local reduction in the workpiece occurs according to the die geometry. The components undergo a 2-dimensional pressure stress state in the reducing zone. This acts favorably on the formability, especially in micro forming.

Due to the geometrical scaling effect, i.e. ratio surface/volume increasing with smaller components, the vulnerability to tensile load also increases, see also 4.3.3. An example is given in Fig. 4.4 (top) showing a swaged sample of spray-formed AlSi10Zn13Cu4 without any detectable defects. The material cannot be formed by drawing due to cracking. Figure 4.4 (bottom) demonstrates the forming capabilities of rotary swaging. The component is a prototype for a micro valve housing formed over a mandrel. With this a calibrated inner diameter can be achieved, which would require an additional machining operation for other forming technologies like cold forging (extrusion).

Important parameters that influence rotary swaging are the total and incremental deformation degrees, the forces acting on the system and the kinematics of the process that indicate the productivity. Figure 4.5 shows the geometrical relationships.

The deformation degree in the radial direction is described as follows (4.1):

$$\varphi = \ln \frac{d_1^2}{d_0^2}$$

(4.1)

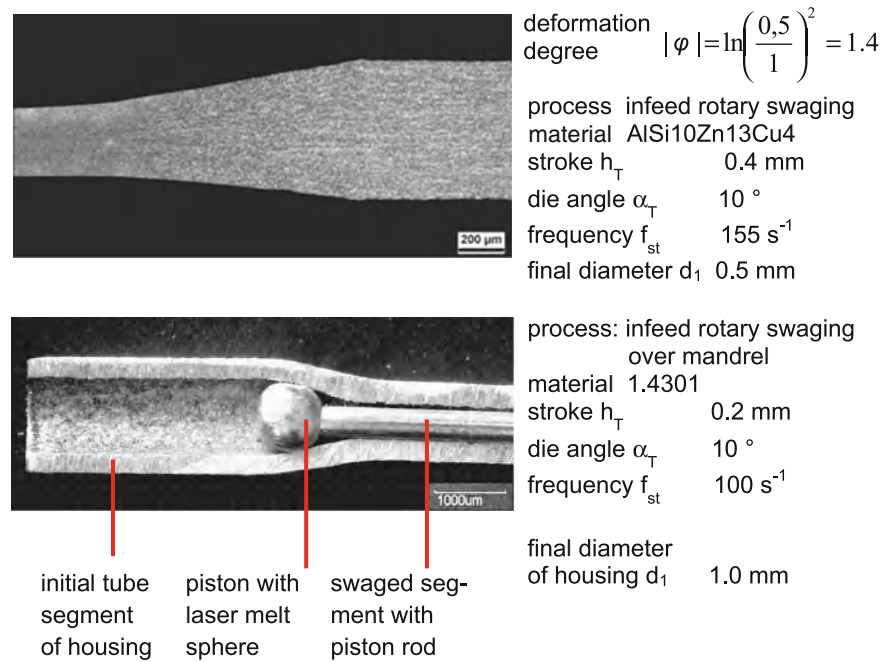
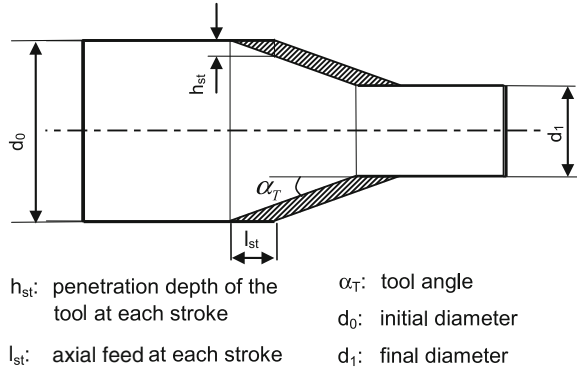


Fig. 4.4 Swaged micro components

Fig. 4.5 Geometrical conditions on the workpiece during infeed rotary swaging



The incremental deformation degree due to each stroke can be defined as:

$$\varphi_{st} = \ln \left(\frac{d_0 - 2h_{st}}{d_0} \right)^2 \quad (4.2)$$

h_{st} is the effective stroke:

$$h_{st} = \frac{v_f}{f_{st}} \tan \alpha_T \quad (4.3)$$

with v_f being the axial feed rate, α_T the die angle, and f_{st} the stroke frequency.

The maximum axial feed rate is obtained from the kinematics as:

$$V_{f,max} = \frac{h_T}{(t_{cy} - t_0) \tan \alpha_T} \quad (4.4)$$

with t_{cy} being the period ($=1/f_{st}$) and t_0 the time segments with dies closed, and h_T the stroke of the base jaws.

Equation (4.4) gives an estimate for the processing time t_{pr} and thus the productivity. Calculating t_{pr} for the micro valve part of Fig. 4.4 would give 50 ms, the achievable production rate in practice is lowered due to handling times and the reduced feed rates because of process limits.

Figure 4.6 illustrates the forces during infeed swaging. Additionally 3 different sections are marked within the dies: reducing zone (I), calibration zone (II) and die outlet (III). The total radial forming force F_F is divided into components in zones I and II, and the related friction components are given as F_{RI} and F_{RII} . The axial forces are characterized by F_F and F_{rej} (rejection force). The neutral surface NS separates the material flow in and against the feed direction.

The rejection force widely influences the process window for rotary swaging. In macro swaging technologies the reducing zone I is roughened by hardfacing. In micro rotary swaging rejection limits the achievable axial feed rate, see Fig. 4.7.

The maximum applicable feed rates that were experimentally found for wire material are about 10 % of the theoretical value given by Eq. (4.4). The findings

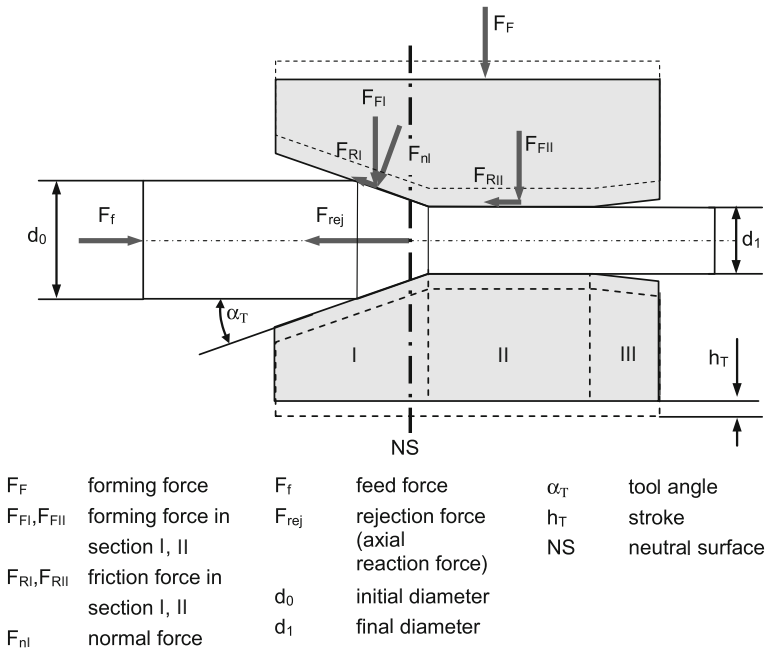


Fig. 4.6 Process forces in infeed swaging

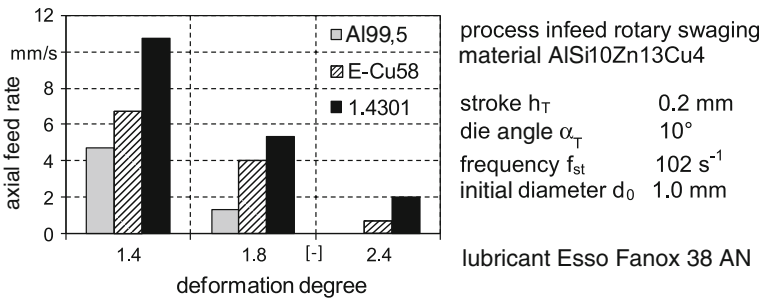


Fig. 4.7 Maximum feed rate in infeed swaging

indicate a strong dependency on the Young's modulus E , which is true for all degrees of deformation. Considering the ratio of part diameter to part length, which is less compared to the macro range, the limiting factor in infeed swaging is Euler buckling.

The buckling force F_K with the axial forces of Fig. 4.5 during the closing time of the dies is (Fig. 4.8):

$$\begin{aligned}
 F_K &= F_f + F_{rej} = 2\pi^2 \frac{EI}{L^2} \quad (a) \\
 F_K &= F_f + F_{rej} = 4\pi^2 \frac{EI}{L^2} \quad (b)
 \end{aligned}
 \tag{4.5}$$

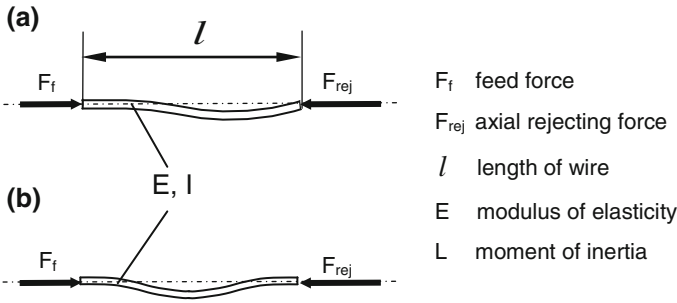


Fig. 4.8 Euler buckling of micro parts

The smallest producible part diameter is limited by the osculation ratio of the tools and workpiece, the opening stroke of the tools during forming and the yield point of the material. The osculation ratio directly affects the achievable final geometry (diameter and roundness) and is difficult to predict analytically. The relation of working stroke to the final diameter is limited. Beyond this limit the workpiece can enter the gap between the tools (Fig. 4.9).

In a 4-segmented swaging head the gap g between the dies in the completely open status is 1.4 times the radial stroke. Design and process-related measures can be taken for the enlargement of the stroke with respect to the work diameter, see also Sect. 8.1.

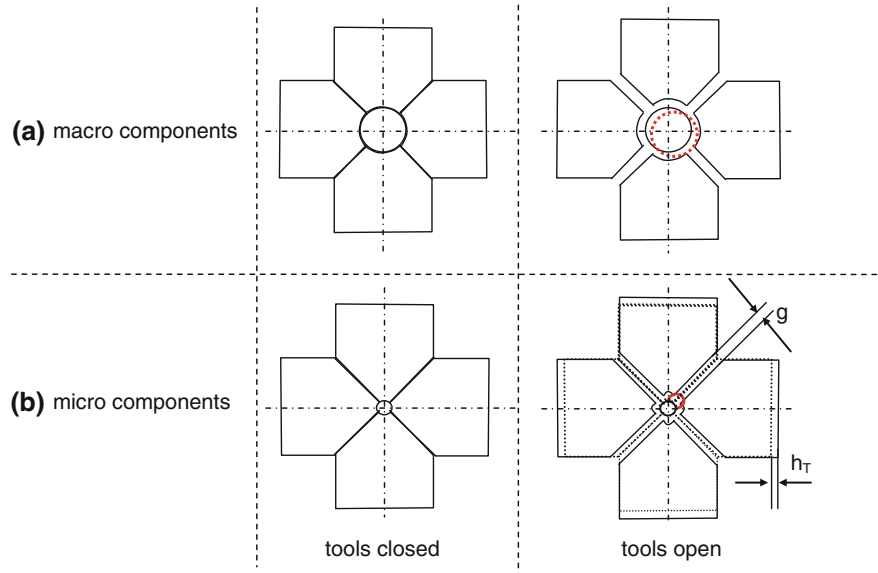


Fig. 4.9 Contact relationship between tool and workpiece

The smallest stroke under which plastic deformation will occur is determined by the yield point. The remaining elastic deformation (spring back) is considered by a smaller contour in the dies than the final work diameter desired.

Besides buckling due to the minor relative axial stiffness of micro parts, other process failures like surface defects, torsion and so-called wing forming (material flow into the gaps of the dies when opened) limit the applicable process windows [15].

4.1.3 Material Effects

The material influence on the result is both similar and different to some aspects in comparison with the macro range.

Figure 4.10 illustrates the part diameter over part length when varying the feed rate. Measurements begin at the initially formed part end.

At the swaged end, a cone-like shape is formed backward to the forming direction, it seems like material flows over the core or the core is less deformed. The smallest diameter is found in this area. As can be seen, the final diameter steadily increases in the first 25 mm of swaged parts from 1.4301. This indicates that near this edge the missing supporting effect of adjacent material reduces the forming resistance. Another effect is obvious, i.e. the final diameter increases with feed rate. This can be explained by the declining number of impacts per volume element and an increase in spring back.

The Martens hardness distribution on steel 1.4301 and aluminum Al 99.5 is shown in Figs. 4.11 and 4.12 respectively.

As expected, hardness after cold forming was higher than in the initial materials, where the hardness is relatively homogenous, with values between 2,400 and 2,750 N/mm². Feed rates of wire have no significant influence on the absolute value of hardness, e.g. hardness values are similar at feed rates of 1.34 and

Fig. 4.10 Final diameters over part length for 1.4301

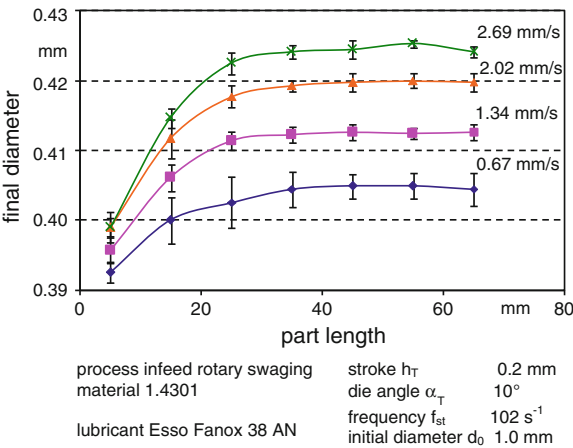
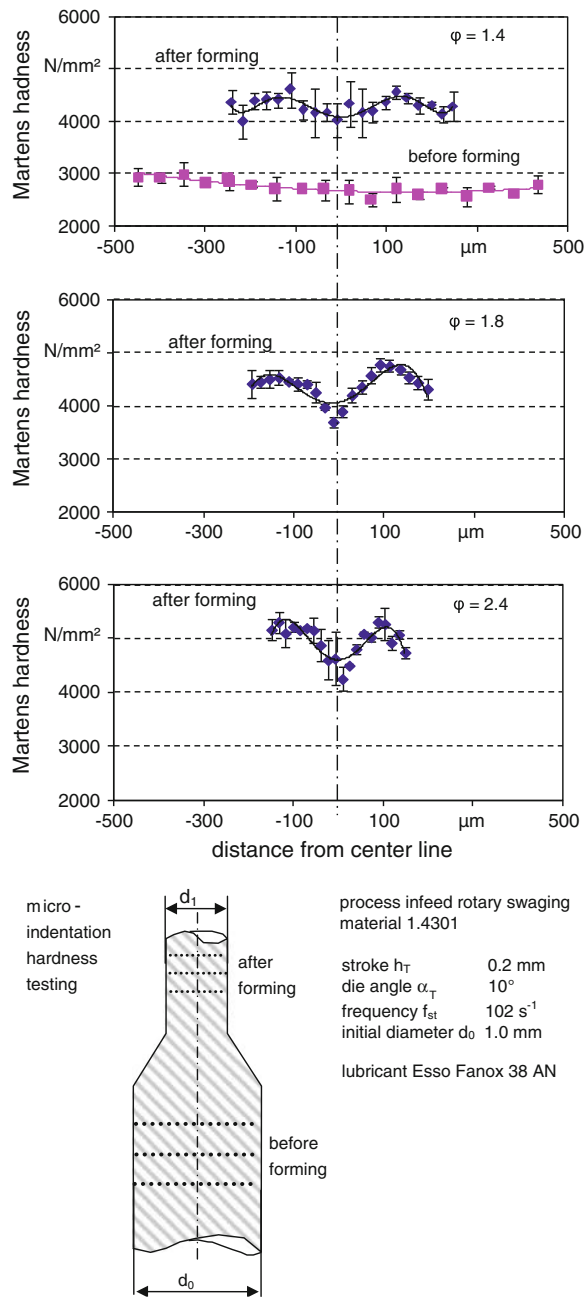
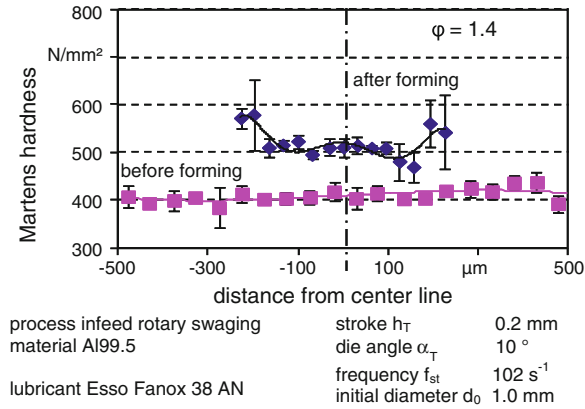


Fig. 4.11 Hardness distributions of 1.4301 steel



10.75 mm/s. The only difference is that the distribution is more homogenous at higher feed rates. An increase in hardness with deformation degree is obvious (Fig. 4.11). For deformation degrees of 1.8 and 2.4 the hardness is lower near the

Fig. 4.12 Hardness distributions of Al 99.5



surface and becomes higher beneath it. An explanation for this characteristic is the Bauschinger effect, namely the consideration of friction acting on the surface.

Material Al 99.5 shows a different behavior. At the studied deformation degree there is also a noticeable increase in hardness from the blank status (420 N/mm²). The increase reaches it maximum values at the surface or very close to it.

4.2 Upsetting

Heiko Brüning

4.2.1 Basics in Upsetting

Upsetting is one of the most important forming techniques belonging to the group of bulk metal forming processes [3]. As stated in DIN 8583, upsetting is a subcategory of free forming and is further divided into flat coining and partial upsetting. This paragraph is addressed to (partial) upsetting as this is the process which is more often applied and has greater practical relevance.

The workpiece is placed between two coplanar surfaces, called tools, which move toward each other reducing their relative distance, thus forming the workpiece and decrementing its height. Due to constant density of the workpiece material, the width of the workpiece increases so that the total volume remains constant. Figure 4.13 schematically shows an upsetting process with friction coefficient $\mu > 0$ between the tools and workpiece.

Depending on the value of friction coefficient μ different behavior of formability is observed. If upsetting is carried out with no or negligible friction between the workpiece and tools, the shape of the workpiece remains as is, that means that the lateral surface stays flat. This can be achieved by either using surfaces well-lubricated with paraffin [18] or by using specially-formed specimens with concave

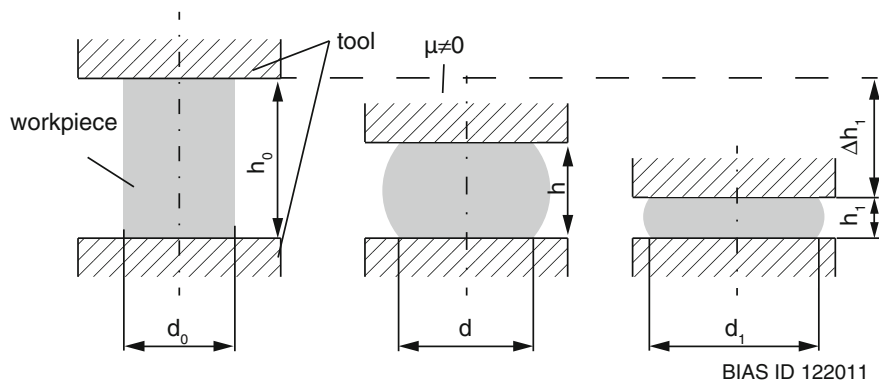


Fig. 4.13 Upsetting of workpiece with $\mu \neq 0$

surfaces, as proposed by [26]. In this case, a homogeneous strain of the workpiece and its grain structure is assured thus leading to a uniform allocation of hardness. For friction coefficients $\mu > 0$ between the workpiece and tools, the lateral surface adopts a convex shape so that the strain is inhomogeneous. In this case the workpiece does not have the same overall hardness, if the hardness is dependent on the absolute value of strain such as for work hardening materials. Figure 4.14 shows a schematical cross-sectional view of a cylindrical specimen which has been processed by upsetting.

The upsetting process is relevant for both theoretical investigations and practical applications. On the one hand, upsetting is a model process to carry out compression tests (see 3.3.1.) in order to determine flow curves, and on the other

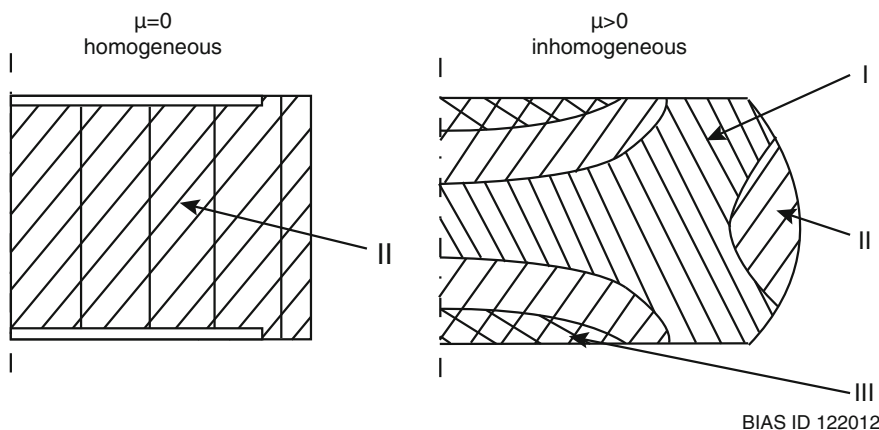


Fig. 4.14 Effect of upsetting of work hardening material on hardness. Area I largest strain, highest hardness. Area II medium strain, medium hardness. Area III lowest strain, lowest hardness

hand upsetting is used daily a multitude of times in many process chains to produce screws, nuts, bolts, rivets and nails. In contrast with machining operations, upsetting benefits from the economic utilization of materials, low cycle times and, one factor which is very important, a homogeneous grain structure making goods mechanically more resistant [3].

There are two main parameters to characterize an upsetting process and its limits:

1. The maximum natural strain φ . This value reaches its maximum, called major strain, parallel to the tool movement direction:

$$\varphi_{max} = \varphi_h = \ln \frac{h_1}{h_0}. \quad (4.6)$$

As soon as a certain value of φ is exceeded, the limit of formability of the workpiece is reached and cracks start to develop.

2. The upset ratio u as a limit for buckling. Buckling will occur when upset ratios greater than u_{max} are processed

$$u = \frac{h_0}{d_0}. \quad (4.7)$$

Exact values determined by experiments for both u and φ can be found in paragraphs 4.2.2 and 4.2.3.

Cold forming of metallic materials generally does not lead to a significant change in density. In conjunction with the fact that material cohesion is not reduced by forming operations, the volume of the workpiece stays constant. This relation can be stated as follows:

$$\varphi_h + \varphi_r + \varphi_t = 0 \quad (4.8)$$

which describes that the sum of the strains in each main direction necessarily equals zero. Among others, such as for compression tests, axially symmetrical cylindrical specimens are used with the consequence that, with equal boundary conditions applied, radial strain equals tangential strain so that:

$$\varphi_r = \varphi_h$$

and

$$\varphi_h + 2\varphi_r = 0. \quad (4.9)$$

A constant strain is achieved as soon as the stress σ in one direction reaches the flow stress k_f . The maximum strain is realized parallel to the direction of the tool movement so that the stress state reaches its major value in the z-direction, called σ_z , at the contact areas between the tool and workpiece. σ_z is dependent on flow stress k_f , the radial distance r to the neutral surface, the upset ratio u of the workpiece and the friction coefficient μ between the workpiece and tools [3]. If a homogeneous grain structure is assumed and the average grain size L_K is much

smaller than the workpiece diameter, σ_z can be calculated analytically by an exponential function [27] as follows:

$$\sigma_z = -k_f * \exp\left[\frac{2\mu}{h}\left(\frac{d}{2} - r\right)\right]. \quad (4.10)$$

For practical applications a simplification of 4.5 is often used. The exponential fraction can be approximated by a series expansion, neglecting terms of higher order as shown in (4.11).

$$\sigma_z = -k_f \left[1 + \frac{2\mu}{h}\left(\frac{d}{2} - r\right)\right]. \quad (4.11)$$

Calculating the maximum tension using (4.10) or (4.11), it is obvious that $\sigma_z = -k_f$ is constant and independent of r if the friction coefficient $\mu = 0$. Figure 4.15 shows all three stress states: upsetting without friction, simplified model with friction, and analytical model with friction.

The required punch force F_z can be calculated by the integration of (4.11) across the surface A of the workpiece. For cylindrical workpieces F_z results in:

$$F_z = -A * k_f \left[1 + \frac{1}{3}\mu\frac{d}{h}\right]. \quad (4.12)$$

4.2.2 Buckling in the Upsetting Process

Buckling is one sort of defect that might occur during the upsetting process. As soon as buckling appears the grain structure is defective, which is highly visible

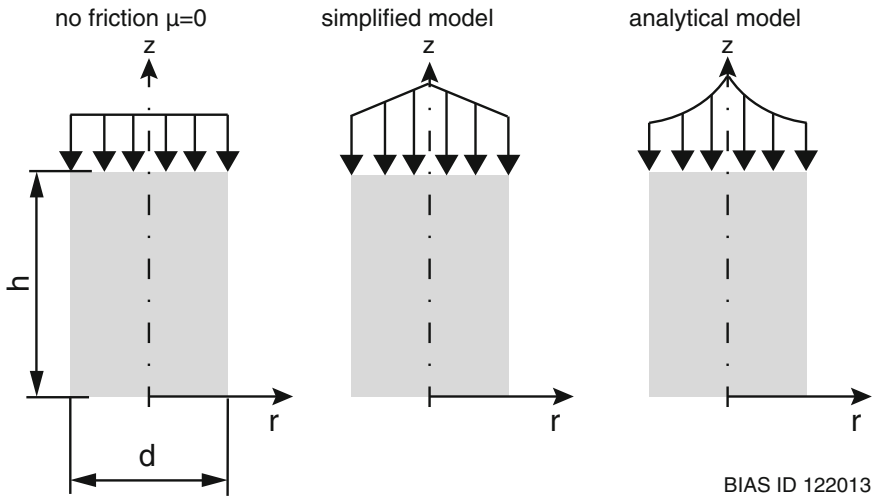


Fig. 4.15 Models of stress curve during upsetting

in cross-sectional polishes with the result that mechanical properties are shortened. Experiments have been carried out by Messner [19] with specimens in the micro range as well as macro range to determine the characteristics of the buckling effect in workpieces with major dimensions smaller than 1 mm. Specimens of brass (CuZn15) as well as an austenitic chromium stainless steel (X4CrNi18-10) have been used. The grain size L_K is adjusted by heat treatment so that its effect on buckling could also be determined. It is to be stated that there is an increasing influence of grain size on the evenness of the lateral surface after the upsetting process for decreasing ratios of sample diameter to grain size d_0/L_K . Specimens with an initial diameter of $d_0 = 0.5$ mm and a grain size $L_K = 65$ μm showed an inhomogeneous lateral surface after the upsetting process, because some crystal grains extruded the surface. Inhomogeneities are not only limited to grains corresponding to the surface but also within the specimen. There are areas whose deformation is larger than others, even though friction between the workpiece and tools is close to zero. The strength of this effect increases with the decreasing dimensions of the workpiece.

Buckling of the specimen occurs as soon as an asymmetrical movement of the material flow perpendicular to the tool movement direction is observed. The value of buckling Δx is measured as shown in Fig. 4.16.

The size of relative buckling $\Delta x/d_0$ is investigated by means of the size of the specimen, the upset ratio and the grain size. As shown in Fig. 4.17, a decreasing size of specimen inevitably leads to an increase in relative buckling, for example an upset ratio $u = 2.4$ for specimens $d_0 = 4.8$ mm causes a relative buckling of 0.05 whereas, specimens with $d_0 = 0.5$ mm experience relative buckling of 0.25. Figure 4.17 only shows these effects for an absolute value of natural strain $|\varphi| = 0.8$, but it is to be noted that similar tendencies are recognized for $|\varphi| = 0.2$ and $|\varphi| = 0.4$. Overall there is no influence of the grain size L_K on relative buckling, not even if the cross-section of the specimen consists of an average of five grains. This effect is valid for both the macro and micro range, and is supported by finite element simulations by Mori et al. [21] for specimens with conventional dimensions. Vollertsen [28] states that the statistical spread increases with decreasing sample size. This can also be seen in Fig. 4.17, as the spread in size of relative buckling is larger for samples with $d_0 = 0.5$ mm compared to samples with $d_0 = 4.8$ mm.

As the micro structure, represented by the grain size, has no significant influence on relative buckling, this size effect can only be based on the shape inaccuracy of specimens due to the machining process. This means that buckling only occurs if the cylindrical specimen or the tool has at least minor deviations in shape compared to its ideal geometry, or if the tool movement is not exactly perpendicular to the workpiece surface. In Fig. 4.18 possible causes for buckling while upsetting are shown. Due to the fact that impurities in the ambient atmosphere such as dust belong to the same order of magnitude as the specimens being upset, it is advisable to carry out micro cold forming operations such as upsetting in clean room conditions.

In the macro range a limit of upset ratio $u = 2.3$ is generally recognized [18]. Due to the fact that the value of relative buckling is a steady function of upset ratio,

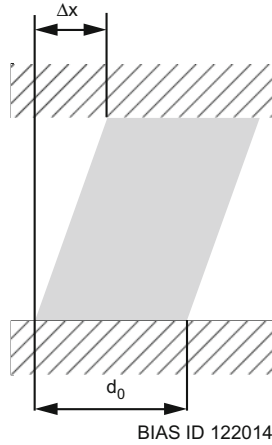


Fig. 4.16 Measurement method for buckling Δx

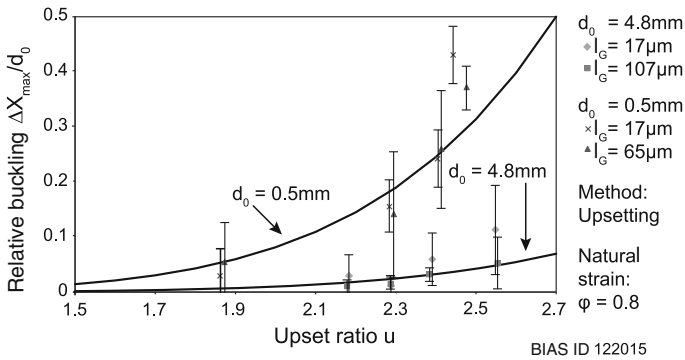


Fig. 4.17 Influence of upset ratio u on relative buckling during the non-lubricated upsetting of cylindrical specimens CuZn15 according to [19]

it is also clear that for ratios $u < 2.3$ upsetting operations lead to buckling and thus to deviations in the shape of specimens. The value of this deviation is accepted for industrial applications. If the value of relative buckling in the macro range is also applied in the micro range, the maximum upset ratio u is smaller, because relative buckling increases with decreasing sample diameter for constant upset ratios. Thus in the micro range the tolerable upset ratio u is far below 2 [19].

4.2.3 Formability in the Upsetting Process

Formability is a value given to describe the capability of enduring the true strain of a workpiece. The limit of formability is reached as soon as parts of the workpiece

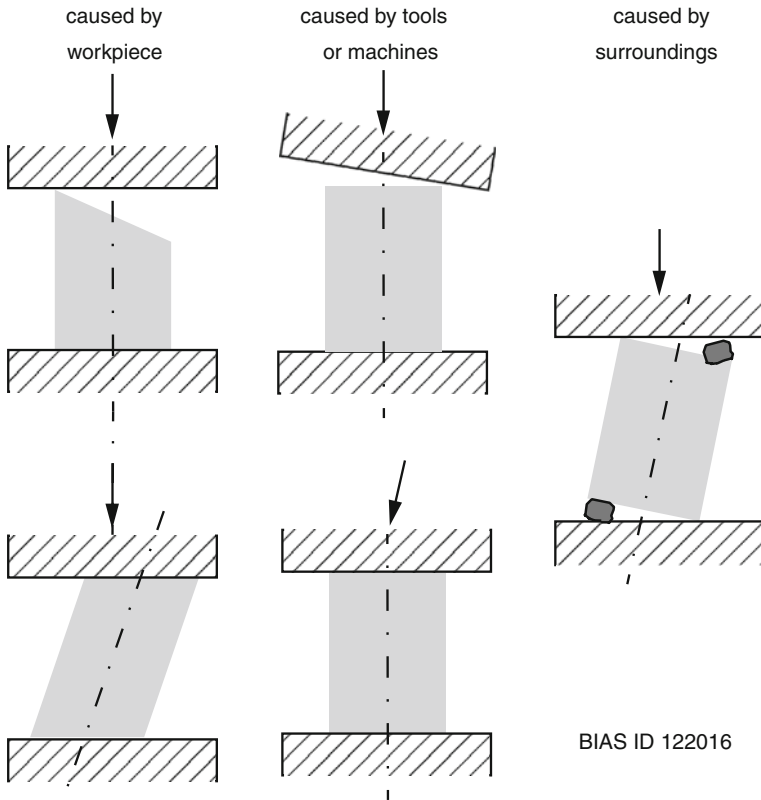


Fig. 4.18 Causes for buckling during the upsetting process

lose their cohesion. An indication of this is the appearance of cracks, which can either occur along the outer surface of the workpiece or inside. Cracks inside the workpiece are easily detectable by analyzing cross-sectional polishes, but also during the upsetting process a discontinuous increase in punch force can be an indication of material failure due to excess formability [12].

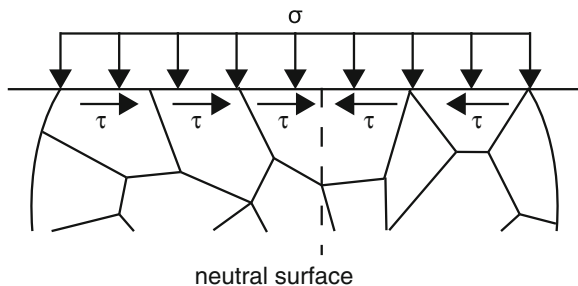
In meso and macro scale, upsetting without any externally detectable defects is possible up to absolute values of natural strain of $|\varphi| = 2.0$ [19]. This value is independent of sample diameter to grain size ratios for at least $d_0/L_K \leq 70$. Nevertheless, a correlation between grain size, surface roughness and natural strain has been reported: surface roughness increases with both increasing absolute values of natural strain and grain size. As shown in Fig. 4.14, the shape of the cylindrical specimen is dependent on the friction coefficient μ between the tool and workpiece surface. As soon as the friction coefficient $\mu > 0$ is obtained, barreling of the specimen takes place. In the macro range, plastic material flow forces grains which used to form the lateral surface of the specimen to move radially and axially, thus forming the top and bottom surface and coming in direct contact with the tool surface.

In the micro range different behavior of specimens with regard to formability is reported. Upsetting is only possible up to $|\varphi| = 2.0$ if $d_0/L_K \approx 30$ or above, which means that a fine microstructure is required. But also in this case, the fillet from the bottom and top surfaces to the lateral surface is very rough, leading to stress peaks. The fact that still no cracks appear indicates that the stress peaks are diminished by locally extended plastic material flow. If the ratio of the sample diameter to grain size d_0/L_K decreases, the maximum natural strain until no cracks appear also decreases. For $d_0/L_K \approx 8$ and a sample diameter $d_0 = 0.5$ mm, at the lateral surface as well as at the bottom and top surface cracks tend to partially develop, starting at $|\varphi| = 1.5$. As soon as $|\varphi| \geq 1.7$ is reached only defective workpieces with cracks mainly located at the top and bottom surfaces can be produced by upsetting. These cracks are oriented tangentially, leading to the conclusion that the maximum formability of the material is exceeded, because neither at the bottom nor at the top surface do radially tensile stresses occur if $\mu > 0$. A further difference between upsetting in the macro and micro range is the fact that the movement of grains which formed the lateral surface before the upsetting operation only come in contact with the tools for larger values of absolute natural strain compared to the macro range.

4.2.4 Influence of Miniaturization on Friction Coefficient in the Upsetting Process

The upsetting process reduces the height of the workpiece. Due to its constant volume the width increases so that there is relative movement between the tools and workpiece perpendicular to the tool movement direction. The value of relative movement is mainly influenced by the friction coefficient μ between the tools and workpiece, as friction elevates the resistance against relative movement [29]. Figure 4.19 shows a schematic cross-sectional view of a specimen and the

Fig. 4.19 Schematic cross-sectional view of specimen with applied load and resulting stress state



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resulting stress state under applied load. The compression stress σ_n is caused by the punch force. The relative movement between the top surface of the specimen and tool surface is restricted in the radial direction by friction, thus leading to tangential stresses τ_F so that a multiaxial stress state exists. The force available for reducing the height of the specimen is reduced because a fraction is needed to transcend friction. So the greater the friction coefficient μ , the greater are both the barreling effect and punch force.

When a forming operation is carried out with rigid tools, the surface deformation of tools can be assumed to be neglected as the surface of the tools has a greater hardness than the workpiece. Due to the fact that the friction coefficient is a consequence of the coaction of the topography of the interacting surfaces, the friction coefficient changes with the increasing strain of the workpiece and its surface [29]. The surface roughness decreases with increasing absolute values of natural strain, because a flattening effect takes place which minimizes the size of the peaks on the surface.

The amount of friction can be influenced by lubricating the interacting surfaces. This can either be achieved by applying suitable fluent lubricant such as oil or paraffin or by coating the tool surfaces with e.g. PTFE, as proposed by [19]. For non-lubricated surfaces there is no size effect reported for the upsetting of work hardening steel X4-CrNi18-10 concerning the size of the specimen or grain size L_K , which means that neither the grain size nor the absolute size of the specimen has an influence on the average value of friction coefficient. But, as also stated by [19], the mean variation in friction coefficient increases with decreasing sample size. For other materials these influences are still under investigation, and inconsistent results are reported. For upsetting with lubricated interacting surfaces, the influence of grain size on friction coefficient is also at the most very small. In contrast with dry surfaces, a correlation between sample size and friction coefficient exists: the smaller the specimen, the larger the friction coefficient. This phenomenon is based on the fact that with decreasing sample size the fraction of closed lubricant pockets is lowered (see 2.2.1), thus increasing the total true contact area between the tool and surface of the specimen, so that the contact forces grow.

One possibility to decrease both the friction coefficient μ and flow stress k_f , is to apply high-frequency vibration to the specimen. The flow stress is lowered by acoustic softening and stress superposition. The advantages are that on the one hand there are lower mechanical forces on the tools and on the other hand there is an improvement in surface quality of the surface of the workpiece into which the vibration is conveyed [31]. In this case the very top of the surface is heated due to friction by the incremental longitudinal movement of the punch, thus enabling a plastic deformation of the roughness peaks so that the surface quality significantly increases.

4.3 Cold Forging

Volker Piwek

4.3.1 Methods of Cold Forging

Cold forging is a process which is not assisted by any heat from the outside, and characterized by the fact that the process starts at room temperature (20 °C) [13]. All methods of cold forging are characterized by high productivity and are therefore suitable for bulk production. Among upsetting and tube drawing, a large area of cold forging is covered by the various extrusion processes where the material is formed due to a relative movement between the punch and the extrusion tool (sleeve) [17]. Figure 4.20 shows the classification of the different variants of extrusion methods with respect to the specific material flow.

The movement of the punch and the material flow by direct extrusion are generally oriented in the same direction. In contrast, in backward extrusion the material is usually deflected by 180°. With the lateral extrusion process the material flow performs a change of direction of around 90°. The 3 methods described are based on a direct transmission of the punch movement to the face of the workpiece blank. With hydrostatic forward extrusion processes however there is no direct contact between the punch and the workpiece. The workpiece is partially embedded in a fluid which is also used for force transmission. In Fig. 4.21 the characteristic geometries for extruded parts are presented according to the above-described process variants respective of their combinations or different process chains [24].

Basically, all ductile materials are formable by extrusion. The main advantages of this resource-efficient process are:

- optimum use of material with uninterrupted grain flow,
- very high productivity and short cycle times,
- high dimensional and shape accuracy,
- high surface quality,
- process-specific cold work hardening.

The convenient material flow (see also Fig. 4.22) and process-specific cold work hardening offers a sufficient strength and stiffness of the extruded workpiece. A near net-shape geometry is very important because of the complicated handling of micro parts. Thus a finishing operation in following process steps can hardly be realized. Here cold extrusion conforms to the high requirements in view of the shape accuracy and surface quality, and is very cost-effective due to its short cycle times. On the other hand, it has to be considered that distortion can occur due to the influence of the friction conditions and the applied material characteristics (e.g. grain structure). Furthermore, the design of the micro extrusion tools has a significant influence on process stability.

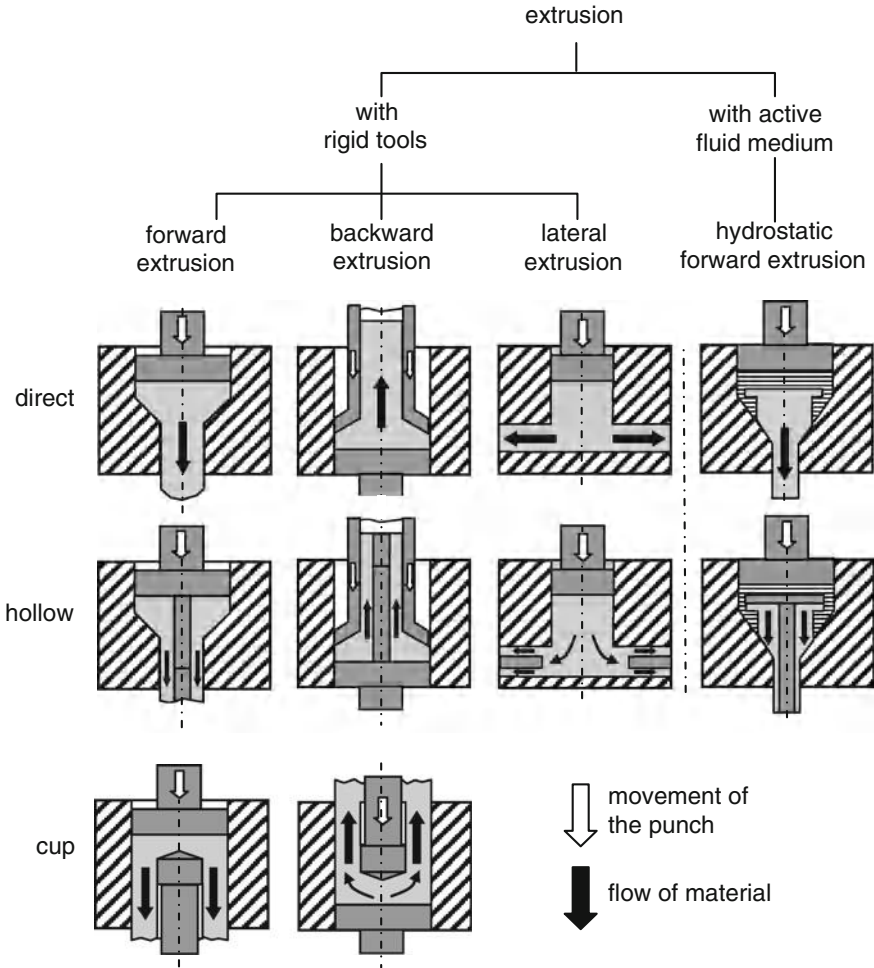


Fig. 4.20 Classification of extrusion processes

4.3.1.1 Fundamentals of Extrusion

Basically forging processes can be stationary or transient. Due to the coincidence of flow lines and trajectories, extrusion processes can be determined as stationary (Fig. 4.22). For a stationary process the machine work W_m is the product of the punch force F_p and the punch stroke h_p .

$$W_m = \int_0^{s_p} F_p ds \approx F_p \cdot h_p \quad (4.13)$$

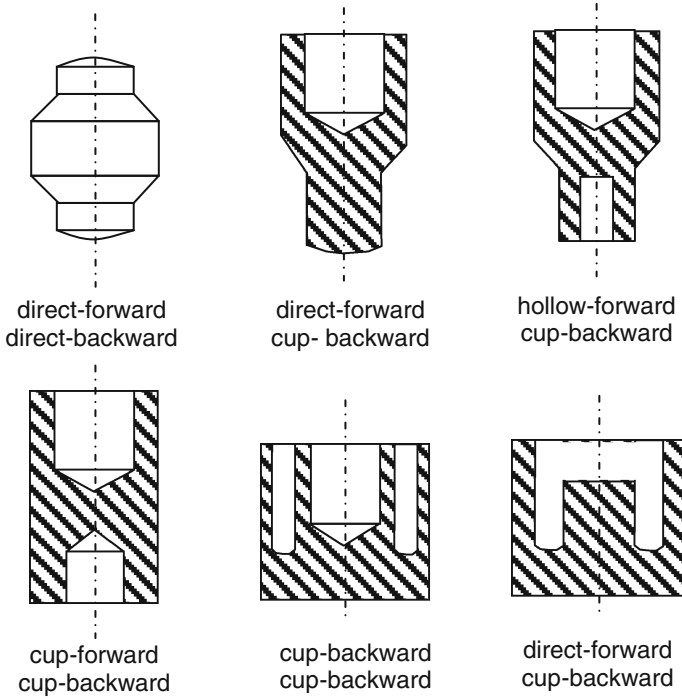


Fig. 4.21 Examples of producible geometries and process combinations

The medium flow stress k_{fm} can be obtained by the integration over the deformation degree φ in the limits from the start (Index 0) to the end of the process (Index 1).

$$k_{fm} = \frac{1}{\varphi} \int_{\varphi_0}^{\varphi_1} k_f d\varphi \quad (4.14)$$

For ideal work W_{id} , considering only the formed volume V of the material and the variance in the effective strain $\Delta\varphi_v$, follows:

$$W_{id} = k_{fm} \cdot \Delta\varphi_v \cdot V \quad (4.15)$$

In practice, the provided punch force F_p has to exceed the ideal work W_{id} due to friction and shift work. These additional losses during the deformation process are detected by the efficiency of deformation η_F . For the required work of the plunger W_{pl} :

$$W_{pl} = \frac{1}{\eta_F} \cdot W_{id} \quad (4.16)$$

Substituting Eq. 4.15 into Eq. 4.16 and equating this term with Eq. 4.13 results in:

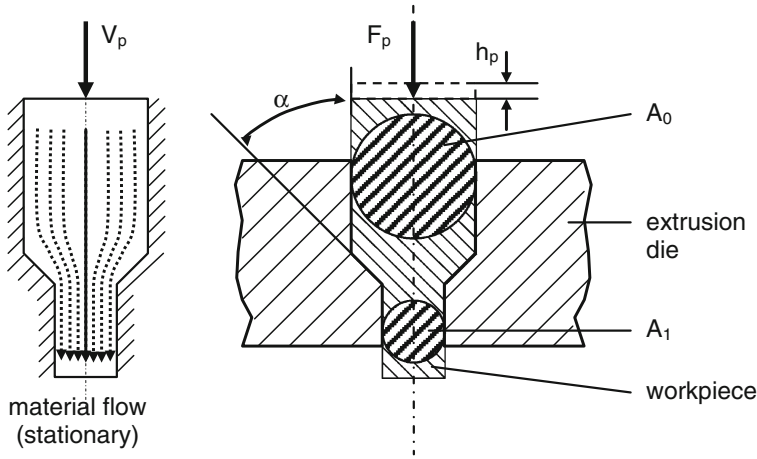


Fig. 4.22 Flow of material (*left*) and schematic diagram of a direct-forward extrusion process (*right*)

$$\begin{aligned}
 F_p &= \frac{1}{\eta_F} k_{fm} \cdot \Delta\varphi_v \cdot V \cdot \frac{1}{h_p} \\
 &= \frac{1}{\eta_F} k_{fm} \cdot \Delta\varphi_v \cdot A_1
 \end{aligned}
 \quad (4.17)$$

In this expression, A_1 is the cross-sectional area after forming (A_0 before forming). The forming forces F_p are increasing with a higher deformation degree $\varphi = \ln(A_1/A_0)$ with respect to the opening angle 2α of the extrusion tool and with a higher material strength and hardness. Figure 4.22 shows schematically the steady flow of material and the process parameters for direct-forward extrusion.

4.3.1.2 Technology for Cold Forging of Micro Parts

The previously mentioned basic principles and relationships are generally independent of workpiece size. The extrusion process is applied for the bulk production of high quality products in the domains of automotive and machine parts, especially for connecting elements, and in electrical industries [8]. Due to the increasing miniaturization of microelectronic and mechatronic assemblies and the complicated application of other manufacturing methods, e.g. metal cutting at once the requirement to realize even those micro components by extrusion operations is given. While the masses of workpieces in the macro scale vary from a few grams to several kilograms, micro parts have only up to some 100 mg of material that has to be shaped.

With the manufacture of micro parts by extrusion, the friction effects due to the small workpiece dimensions (surface-volume ratio) and the grain size and its distribution exert a significant influence on the entire forming process [6]. The

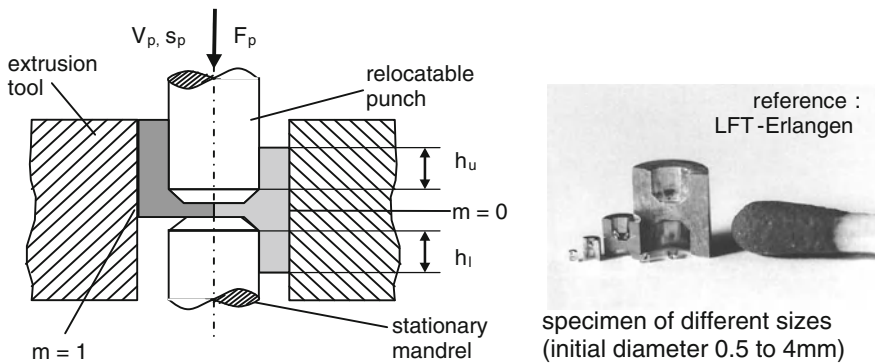


Fig. 4.23 Schematic diagram of the Double-Cup-Extrusion-Test (left), and various sizes of specimen [6] (right)

friction can be up to 20 times higher than in the macro area [4]. On the one hand this influence is reflected in a decreasing efficiency of deformation η_F and on the other an enlarged deviation of the formed part's geometry appears [7]. The most important test for detecting friction conditions during extrusion is the Double-Cup-Extrusion-Test (DCE), wherein a cup is formed in the forward direction whilst at the same time being generated in the reverse direction (Fig. 4.23).

In this test, the ratio of upper to lower cup height h_u/h_l is a measure of friction during the forming process. For the ideal situation, without any friction ($m = 0$), two cups of equal height ($h_u = h_l$) will result, while at maximum friction ($m = 1$) only a cup geometry in the reverse direction of the punch feed will be formed. Therefore the analysis of the tribological effects of design processes for extruding metallic micro parts has a significant relevance [9].

Experimental results of the DCE-Test on CuZn15 specimens with diameters from 4 mm down to 0.5 mm show a significant influence of the specimen size on the friction coefficient [5]. Furthermore, there are other important methods to determine the friction conditions within the field of micro forming. The micro-ring-compression-test, for example was used, by [2] to determine the friction on aluminum alloy 6061 specimens in a diameter range between 2 and 3 mm. Diameter dimensions below 2 mm down to less than 1 mm were explored for copper alloy CuZn30 by [14] using the pin-length-test. In addition to the experimental investigations, analytical or numerical methods, e.g. finite element method (FEM), can also be applied to determine the friction coefficient. Here the determination of the forming forces is based on the Upper-Bound-Method and the slice-model. The displayed results of a FE-analysis are based on a coarse grained structured material with an average grain size of 211 μm . The determined friction coefficients by the different methods and size ranges of the specimen are arranged in Fig. 4.24. Generally the tendency can be recognized that friction increases with the smaller dimensions of the formed test piece.

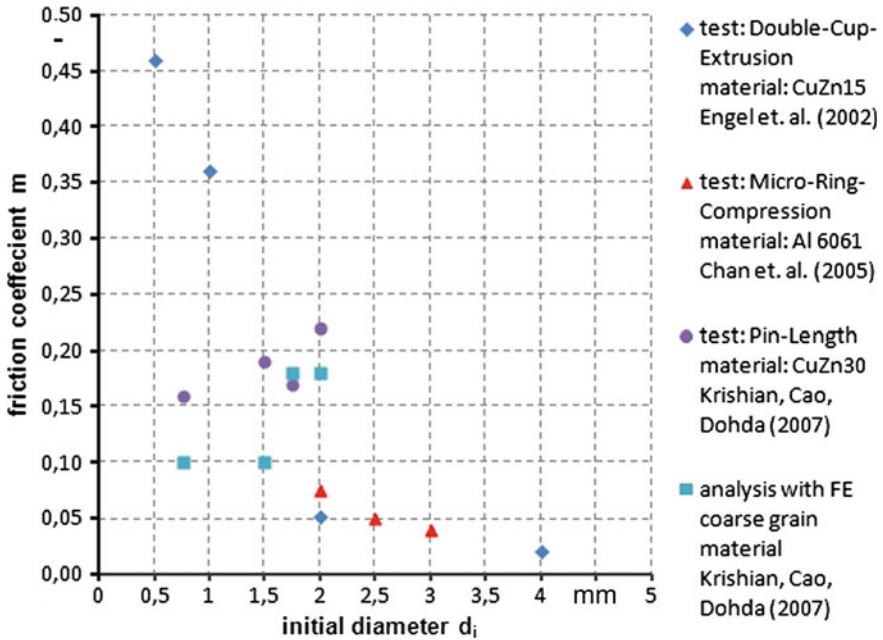


Fig. 4.24 Determined friction coefficients by different methods

On the one hand, the FE approach shows similar results to the Double-Cup-Extrusion-Test in that the length of the formed workpiece section decreases with increasing friction. On the other hand it was found that the friction coefficient again decreases with an increasing workpiece diameter, but distortion effects become significant [20]. These effects can be primarily explained with the grain size effect in the material and not with the friction conditions between the contact surface of the punch and the surface of the specimen. Generally, to reduce friction the use of lubricants is suitable. Furthermore, coatings of the tool surfaces made of silicon-based Diamond-Like Carbon(DLC), chromium nitride (CrN) and titanium nitride (TiN) can be considered.

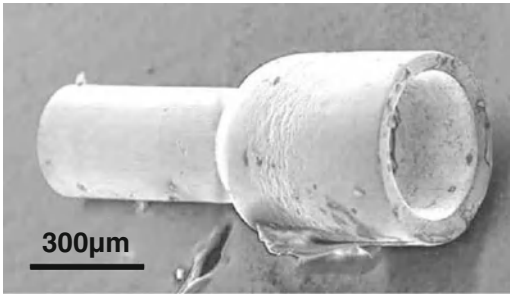
4.3.1.3 Component Applications in the Micro Range

Figure 4.25 shows examples of extruded metallic micro parts for industrial applications (a, c) and the field of research (b).

Workpieces (a) are produced by direct-forward cup extrusion on an industrial extruding press. Component (b) has geometry similar to the shaft of a micro motor and was manufactured by direct-forward in combination with cup-backward extrusion with a shaft diameter of 0.5 mm and a wall thickness of 50 μm . Copper



(a) micro parts produced by extrusion (reference: Wafios GmbH)



(b) direct forward / backward cup extrusion (reference: LFT)



(c) direct forward / backward cup extruded copper pin (reference: Wafios GmbH)

Fig. 4.25 Examples of extruded metallic micro parts

pin (c) was manufactured by the same process combination with a shaft of 0.5 mm and a nominal inner cup diameter of about 1 mm.

4.3.2 Materials: Process Windows and Limits

Steel as well as nonferrous metals are formable by extrusion. While primarily unalloyed and low alloyed steels are used for macro components, such as screws,

piston pins and piston rods [23], aluminum and copper are in particular applied in the production of micro components in the field of electrical engineering. The most important aluminum material for extrusion forming operations is pure aluminum Al 99.5 and for higher strength requirements self-hardening (thermosetting) alloys, e.g. AlMg5, AlMgSi and AlCuMg. Furthermore, titanium-based light metals alloys, especially titanium-aluminum, are suitable for the extrusion process. Pure copper is mainly used in the electrical industry, such as E-Cu, Cu and Se-Sf-Cu, in the production of contact pins for example. In the group of copper alloys, brass materials (α – brass) with a tin content of 28 % up to a maximum of 37 % (CuZn28 ... CuZn37) are particularly suitable. Also bronze materials, especially tin- and silicon-bronzes, e.g. SnBz1 and SnBz2Mn, are used.

For forming operations in the macro range, guidelines for accessible material-related deformation degrees are available. Due to the significant increase in the friction conditions in the micro range, the relevance of these values is limited. From the macro range it is known that the highest deformation degrees can be reached for direct-forward extrusion of pure aluminum with values up to 4. This is also applicable with some restrictions to hollow-forward extrusion. The deformation degrees for copper and bronze materials range from 1.2 to 1.6. However, with heat treatable or stainless steels, in general only deformation degrees of less than 1 can be realized.

Especially for micro forming, the achievable shape accuracy is influenced by the grain size and its distribution (anisotropy) in the blank that is to be extruded, and the friction conditions between the contact surfaces of the workpiece and tools [4]. Extended investigations have shown that the occurring distortion of extruded micro parts is caused by irrational local deformation, spring back effects and interfacial friction stress [2]. These features are size-dependent so that reduced shape accuracy is expected with a decreasing size of specimen. Research on extruded brass pins indicated an influence of the grain size on the distortion achieved. Thus, a material with coarse grains shows a significant straightness deviation, while a fine-grained structure is less affected [14].

4.3.3 Extrusion Tools for Micro Forming

Figure 4.26 illustrates the basic design of a tool for a direct-forward extrusion process with special features for micro forming. The tool consists of an upper, vertically movable section, where the punch is supported on the upper base plate and can be adjusted in the vertical direction by a set screw.

The geometry of the workpiece is produced by the extrusion tool that consists of an inner sleeve, which is strengthened by ring reinforcement. In the vertical direction, the tool is supported by the lower base plate, in which the ejector is also linearly mounted. This element works against the punch direction and enables the detachment of the workpiece after the forming operation. In extrusion tools for micro forming the matrices have correspondingly small holes in a comparatively

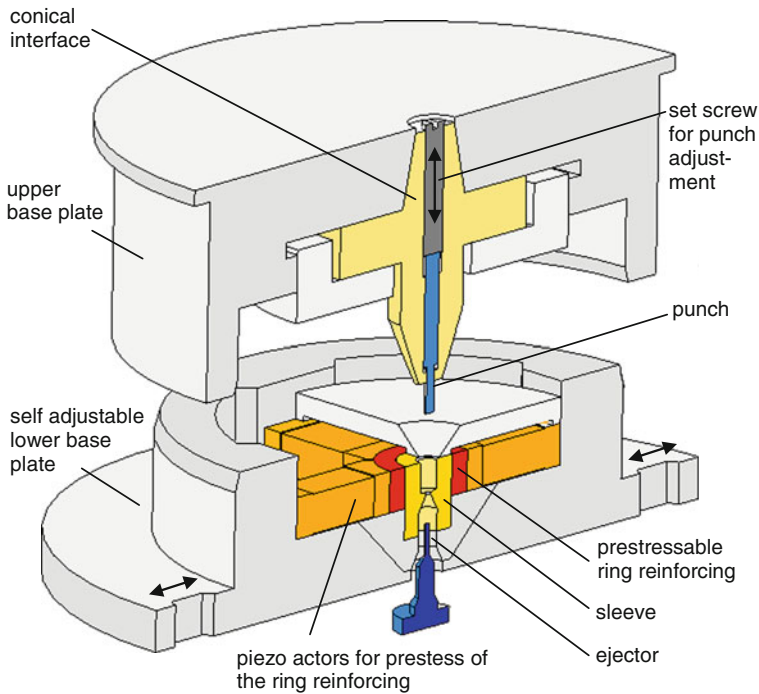


Fig. 4.26 Tool for a direct-forward extrusion process

large sleeve, which is generally not limited in its external dimensions. The punches, or mandrels, are usually very slim structures, where buckling resistance is one of the fundamental design criteria. Due to this requirement in the micro range, the demand for wear resistance for the stationary as well as the relocatable tool components in compression molding increases significantly. The general requirements on extrusion tools are e.g.

- conical design of the interfaces between several parts
- axis symmetrical design of the parts to balance out non-coaxialities.

For design in the micro range there are specified guidelines, such as

- a minimum number of tool elements to minimize the tolerances of the entire tool assembly
- floating respectively self-adjusting positioning of punch and pin in the horizontal direction
- adjustable punching tool respectively reinforcement.

The axially symmetrical and self-adjusting designs of the extrusion sleeve and a coaxial alignment to the punch are both important to fulfill the requirements of close fittings, especially for micro extrusion processes. The conical design of the interfaces eliminates the clearance that is needed for mounting. Furthermore, there

are approaches to optimize the accuracy and the ejection process of the workpiece, which is complicated by the narrow clearances in the micro range, by using a piezo-driven prestress and expandable punching tool [25].

In addition to the basic requirements of high shape accuracy and durability of the tools, in the micro range there are approaches for flexible tool systems to cover an enlarged spectrum of workpiece geometries [30]. The goal of such systems is to provide tools with modular-like structures for the various methods of extrusion or combination thereof [22]. There are essentially three types of tool component: basic-, process- and workpiece-specific components. To reduce the time for changing of the tool (machine setup time), quick-change tooling systems can be used, such as a clip device for the attachment of the punch. Furthermore, the opportunity is given to separately renew or exchange specific areas with increased tool wear due to friction.

Another important feature of the tools is the material used. A suitable material for an extrusion tool has to fulfill on the one hand the requirements of its own manufacturing process as well as for its subsequent application in forming operations. Figure 4.27 shows the main requirements of such materials divided in two groups with focus on the tool manufacturing and the forming process [13].

Generally hardened and tempered tool steels and carbide metals are used, which provide hardness from 50 up to 67 HRC and a bending and burst strength from 2.400 to 2.800 MPa and 3.000 to 4.100 MPa respectively. Furthermore, powder metallurgical steel is used to achieve a high material homogeneity and a fine carbide grain structure (grain size from 2 to 6 μm), while in conventional tool steel

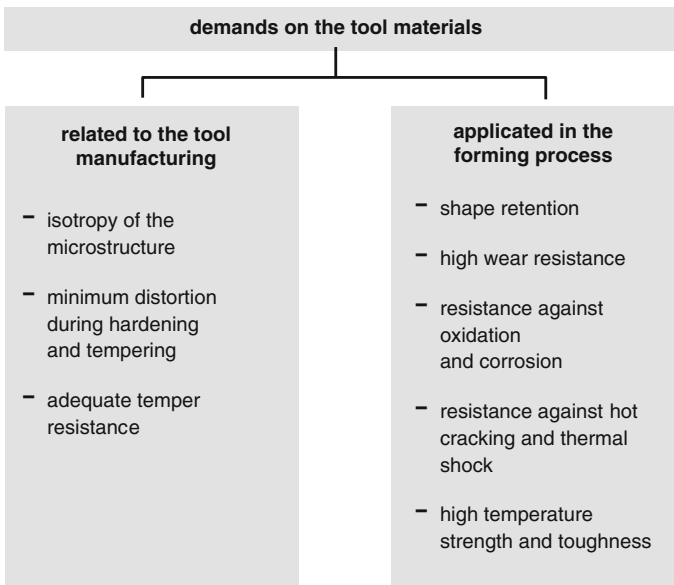


Fig. 4.27 General demands on the materials for extrusion tools

the carbide size is between 30 and 50 μm [1]. In particular, a high bending strength is necessary for punching of micro extrusion tools to achieve a sufficient buckling resistance, due to the small dimensions and the thickness ratio required.

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Chapter 5

Sheet Metal Forming

Frank Vollertsen

Latin

d_0	pre-hole diameter (mm)
d_P	punch diameter (mm)
F_P	punch force (N)
F_{BLH}	blankholder force (N)
g	drawn clearance (mm)
I	Inertia moment ($\text{kg} \times \text{m}^2$)
K	spring back ratio
k_f	flow stress (N/mm^2)
r_D	die radius (μm)
s_0	initial sheet thickness (μm)
s_1	final sheet thickness (μm)

Greek

β	drawing ratio
$\beta_{\max}$	Limit drawing ratio
φ	logarithmic degree of deformation
$\sigma_{pl\ 0.2}$	yield strength (N/mm^2)
σ_t	tangential stress (N/mm^2)
σ_r	radial stress (N/mm^2)
σ_{ax}	axial stress (N/mm^2)

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5.1 Deep Drawing

Gerrit Behrens

5.1.1 Single Step Deep Drawing Processes

5.1.1.1 Deep Drawing Basics

Deep drawing is one of the most important processes in sheet metal forming and has a great application potential for manufacturing parts with complex shapes, even those with very small dimensions. Examples of single step deep drawing parts are shown in Fig. 5.1.

According to the definition in DIN 8584, deep drawing is classified as the tensile compressive forming of a sheet or a foil to a hollow body, open on one side, without an intentional change in sheet thickness. As the blank is drawn into the die cavity by the punch, a three dimensional shape is generated (see Fig. 5.2). The drawing force, necessary for the forming, is transmitted from the punch to the workpiece bottom and afterwards into the forming zone in the flange. Caused by the decrease in the blank diameter during the deep drawing process, tangential stresses lead to a compression of material in the flange area. These tangential stresses may cause wrinkling of the blank material, which is why in most deep drawing processes a blank holder is used to exert a normal force on the flange material and press the blank firmly onto the drawing ring. Consequently, the formation of wrinkles is prevented. The stress due to the blank holder pressure is small compared to the radial and tangential stresses of the process.

Since in micro deep drawing very small blank holder forces are obtained, the measurement and control of these forces during the process is highly complex. In order to avoid this effort and the resulting costs, several applications do not use a force controlled blank holder. Instead, the blank holder is set to a fixed position with a gap between the blank and the blank holder. For instance [33] used a gap of

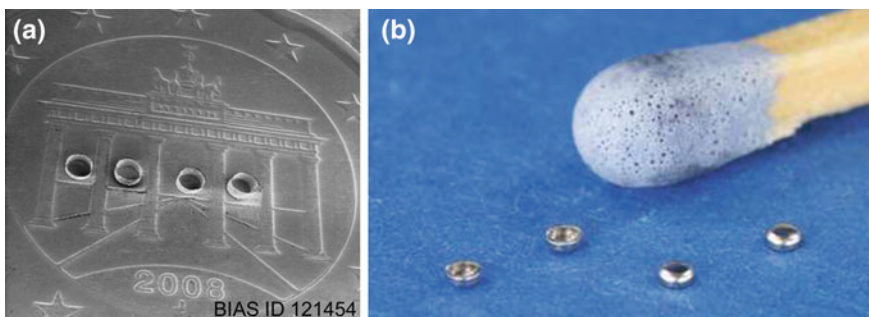
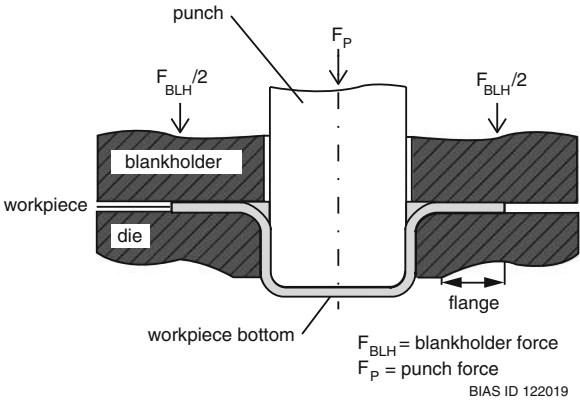


Fig. 5.1 Single step deep drawing parts. **a** BIAS GmbH. **b** Hubert Stüken GmbH & Co. KG

Fig. 5.2 Schematic of the deep drawing process



1/3 of the foil thickness. The gap can be changed using thin plates. Due to this, no precise adjustment of the blank holder force during the process is necessary. However, for every change of a relevant process parameter such as tool or foil geometry, a change of tool set is necessary. In [24] a double-axis micro forming press developed at BIAS is presented, which allows the very precise control of the blank holder force even in the micro range. Therefore, high accuracy force- and stroke-measurement and two independent machine axes, driven by electrical linear motors, are used. If this type of machine is used, optimized for the increased requirements of micro scale manufacturing, a very flexible reaction to changes in process is possible. Changes in tool geometry or foil thickness can easily be adjusted.

The process limits in deep drawing depend on the properties of the sheet material, on the lubrication conditions, on the tool geometry and the forming parameters. In sheet metal forming, the forming behavior is described using two characteristic diagrams. One is the forming limit diagram which is described in detail in Sect. 2.3. The second is the deep drawing diagram, shown in Fig. 5.3. Herein, the initial blank holder pressure is plotted over the drawing ratio. The

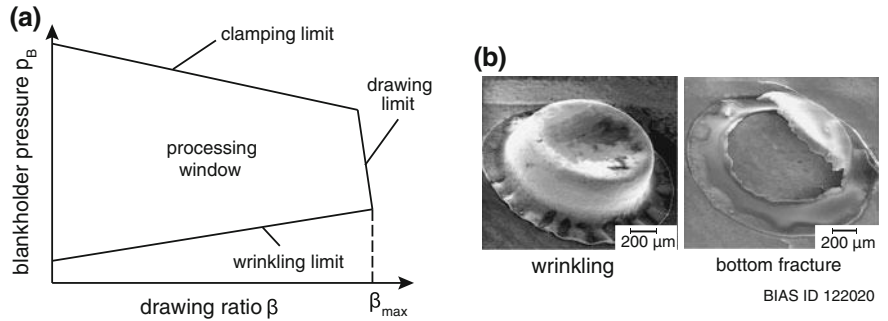


Fig. 5.3 **a** Schematic deep drawing diagram. **b** Failures in micro deep drawing: wrinkles (left) and bottom fracture (right) [24]

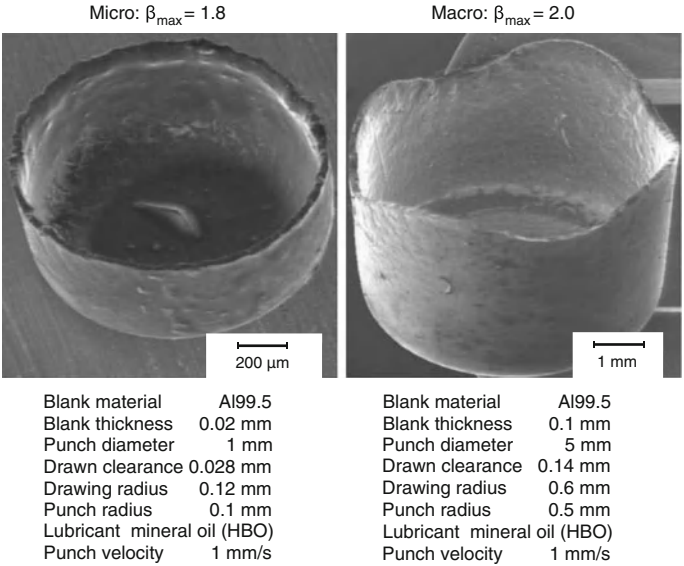
drawing ratio is the ratio between initial blank diameter and punch diameter, it being desirable to achieve high ratio values.

The upper process limit is characterized by component failure at the beginning of the process and is named clamping limit. Above this limit, the part is just stretch formed and fails before the blank can be drawn into the die cavity. Therefore the cup height achieved is very small and a hole is punched through the sheet material. If the values for the initial blank holder pressure are selected below the wrinkling limit, workpieces with wrinkles are obtained (Fig. 5.3b). Only within a certain range will sound parts be acquired. This area is limited to its right side by the drawing limit; exceedance of this drawing ratio results in bottom fractures caused by too strong material hardening in the flange area (Fig. 5.3b). A single value can occur instead of a limit line, and this intersection between the clamping limit and wrinkling limit defines the limiting drawing ratio. This is the maximum value which can be achieved under the given process conditions and is denoted $\beta_{\max}$. The limiting drawing ratio is a characterizing indicator for the deep drawing process, and shows how much the workpiece can be deformed during the process. Higher values of $\beta_{\max}$ indicate improved formability of the material and a better application in industry.

5.1.1.2 Influence of Miniaturization on Deep Drawing

Different investigations into micro deep drawing uniformly revealed a significantly smaller process window and a decreased limiting drawing ratio in comparison with a conventional deep drawing process. For example in [58], deep drawing with punch diameters of 1 and 50 mm were carried out to investigate the size effect on the limiting drawing ratio. The workpiece material was Al99.5 and mineral oil was used as the lubricant. Under the same forming conditions and scaled tool geometry, a limiting drawing ratio of more than 1.8 could be reached in macro deep drawing by the use of the 50 mm punch. In the miniaturized process a limiting drawing ratio of only 1.5 was acquired. A similar study was done by Hu and Vollertsen [24] where a reduction of the punch diameter from 5 to 1 mm resulted in a decrease in the limiting drawing ratio from 2.0 to 1.8 when using the same material (see Fig. 5.4). In [25] the validity of this tendency was also demonstrated for copper blank material.

There are several explanations for decreased drawability in micro deep drawing. One explanation discusses the contribution of the number of grains in the direction of the sheet thickness. If the process is scaled down, the sheet thickness decreases. If the grain size remains constant, the number of grains in the cross-section of the sheet also decreases. Fewer grains mean less grain boundaries, which can result in lower work hardening during forming. In thick sheets which have a polycrystalline structure, the first strained grains will stop straining due to this hardening, while concurrently some grains somewhere else, which are not hardened, begin to strain. This procedure causes a straining of all grains within the forming area. In thin foils, where only one grain covers the sheet thickness, a



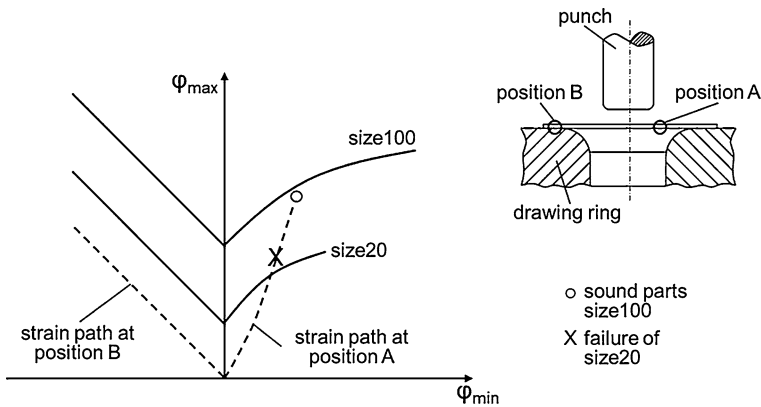
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Fig. 5.4 Reduced limiting drawing ratio in micro deep drawing [24]

Tiffany structure exists (see Sect. 2.3). The orientation of the individual grain determines in which grains the straining starts, and the probability for any existing neighboring grains with similar orientations to continue the straining decreases. Therefore some local grains are strained until the fracture strain is reached and failure of the workpiece occurs. According to that, the amplification of the localized strain in very thin foils leads to less formability and a smaller limiting drawing ratio [25].

Figure 5.5 shows schematically what occurs: At the beginning of a deep drawing process the material is stretched when the punch contacts the blank, as long as the static friction between the sheet and the blank holder and drawing ring has not been overcome. The critical part is the stretch drawn area at position A, which is located between the punch and the drawing ring. According to the weaker deformation behavior of Tiffany structures, the forming limit line is shifted to lower strains in the forming limit diagram as we decrease the size. Due to this, the forming limit in the stretch drawn area at position A is exhausted locally and a fracture sets in earlier than expected compared to experiences in the macro range. So the change in the limiting drawing ratio due to miniaturization can be explained by a density size effect, which leads to a Tiffany structure and corresponding material behavior [59]. The tendency towards decreasing formability, when decreasing the foil thickness, is also shown in Sect. 2.3.2 (Fig. 5.7).

Another reason for a smaller limiting drawing ratio in micro deep drawing is assumed to be the increased relative geometrical deviation in the manufacturing of



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Fig. 5.5 Schematic limiting drawing diagram with size100 material showing continuum behavior and size20 showing Tiffany structure behavior [59]

the tools. As Hu [22] showed, this can result in an increased punch force of more than 10 %, which significantly reduces the limiting drawing ratio.

One of the major factors affecting the deep drawing process is friction at the flange and the radius of the die. Friction is also affected by size effects when transferring the forming technology from macro to micro forming. As shown in Sect. 2.2.1 there is a significant influence of scaling on friction in metal forming when lubricant is used. Generally, an increase in the coefficient of friction with decreasing process dimensions occurs. An explanation for this behavior is given by the lubricant pocket model. As a consequence of increased friction, higher punch forces are reached for micro deep drawing and the limiting drawing ratio becomes smaller. To investigate the tribological size effect for the integration of the real frictional conditions into more realistic models of a micro deep drawing process, e.g. in FEM-simulation, specialized measurement methods for the coefficient of friction in scaled deep drawing have been developed. Further information about these testing procedures and about the cause and effects of the tribological size effect are given in Sect. 2.2.1.

5.1.1.3 Deep Drawing Without Lubricant

Even though the occurrence of the tribological size effect complicates the understanding and determination of the micro forming process, the use of lubricants is dominant in micro deep drawing. This is problematic, since adhesion has a greater influence when downscaling the process and the handling of workpieces becomes more difficult. As micro forming technologies are supposed to be used in mass production processes, it is not desirable to have fabricated parts sticking together. Due to their size the cleaning of products is more complex. Moreover, the

saving of lubricant might help limit harmful effects on the environment and establish healthier working conditions. It is therefore advantageous if the use of lubricants in micro forming is reduced or even avoided.

Since the thin workpiece material in micro forming usually has a lower formability, which leads to a more sensitive dependence of the forming limit on friction between the workpiece and forming tools, it is highly desirable to precisely control the friction conditions. One promising possibility to enable lubricant-free forming is the use of special coatings on the forming tools. This measure is not only aimed at reducing friction, but also to increase the wear resistance of the tools to improve tool life.

Investigations on amorphous DLC coatings as well as on TiN-coatings were carried out in micro deep drawing experiments. Reference [19] investigated the influence of lubrication conditions on the limiting drawing ratio in micro deep drawing using DLC-coatings for blank holder and die. The results show a strong increase of limiting drawing ratio from 1.8 for uncoated tools without lubricant to 2.1 for DLC-coated tools. Lubrication with castor oil has been able to improve the limiting drawing ratio only insignificantly to a value of 1.9. Similar investigations by Hu [26] conducted with DLC and TiN-coatings on dies revealed a significant reduction in the deep drawing punch force for DLC-coatings compared to conventional lubrication with mineral oil (see Fig. 5.6). The TiN-coating did not show such a clear tendency, but might be interesting for dry micro forming because of its low wear factor. Both investigations indicate that coating of the tools shows advantages in micro deep drawing and a great application potential in lubricant-free micro forming.

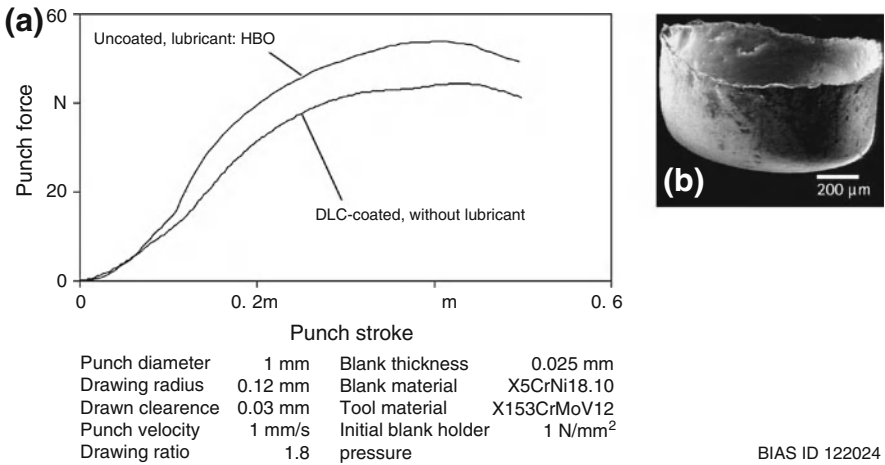


Fig. 5.6 a Comparison of punch versus stroke curves of micro deep drawing with DLC-coated and uncoated forming tools [26]. b Sound part of lubricant-free micro deep drawing [26]

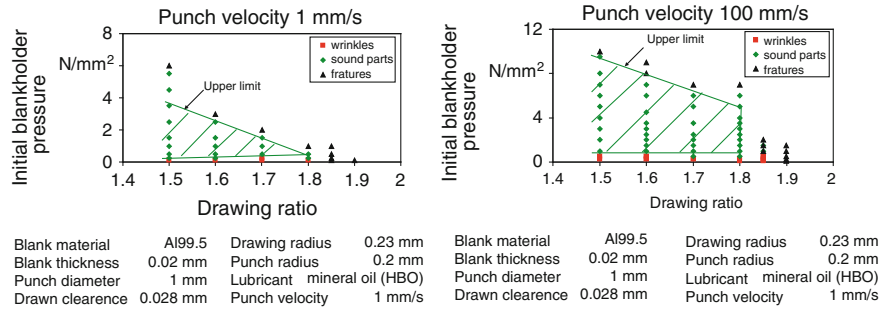


Fig. 5.7 Influence of punch velocity on the deep drawing process window [24]

5.1.1.4 Influence of Punch Velocity

The potential application for micro deep drawing is in the manufacturing of products in industrial mass production. Therefore high punch velocities are required. Conventional micro deep drawing investigations are carried out under laboratory conditions with relatively low punch velocities, like 1 mm/s. Process scaling into the micro range has the advantage of having less moving mass. The reduced weight of the forming tools allows higher cycle rates and punch velocities in micro forming. This is why a good understanding of the process under raised velocity conditions is necessary. Investigations into micro deep drawing with increased punch velocities show an upward shift in the clamping limit in the process window with increasing processing speed for the blank material Al99.5 [24]. This trend is shown in Fig. 5.7 for punch velocities of 1 and 100 mm/s. Thus, the danger of fractures caused by too high a blank holder pressure decreases and the process becomes more stable. However, no changes in the limiting drawing ratio $\beta_{\max}$ could be found for higher punch velocities.

The trend in the enlargement of the process window to higher blank holder pressures can be explained by the dependence of the friction coefficient on the velocity. This behavior is usually described by a Stribeck curve. According to which the area of boundary friction decreases and the area of hydrodynamic friction increases when higher relative speeds occur in lubricated friction. This results in decreased friction between the workpiece and forming tools. Therefore, a higher external load can be transmitted to the workpiece until the fracture limit is reached. Further investigations qualitatively revealed the same tendency towards enlargement of the process window for macro deep drawing processes, though more pronounced than in the micro range. This can be explained by the lubricant pocket model, described in Sect. 2.2.1. Since hydrodynamic friction mainly exists in closed lubricant pockets and their amount decreases with miniaturization, there is less hydrodynamic friction in the micro forming process and consequently the effect is not quite as pronounced. Changes in material behavior as an explanation for the elevated clamping limit with increased velocity should be excluded, since

the maximum strain rate is lower than 150 s^{-1} and at a speed below 200 s^{-1} no change in the flow stress is expected [59].

However, the changing frictional conditions due to velocity variations do not influence the drawing limit. This phenomenon could be explained by the fact that the blank material Al99.5 showed no influence of the strain rate on the forming limit [23]. Therefore, the strain condition is independent of the forming velocity. When the tool geometry and the blank material properties are kept constant, just the drawing ratio determines the occurring maximum strain in the deep drawing process. The change in frictional behavior due to different punch velocities can change the punch force during the process, but it does not influence the forming limit. Thus the processing speed shows no influence on the limiting drawing ratio [23].

Another investigation focusing on the influence of the process speed on the micro deep drawing process was conducted by Justinger [33]. The punch velocity was varied from 0.01 to 100 mm/s using CuZn37 foils. For comparison, the wall thickness distributions on the produced cups were measured using a microscope. The experiments show that the punch velocity only has a small effect on the cup geometry. The cup geometry seems to be more affected by the microstructure.

5.1.1.5 Deep Drawing of Rectangular Parts

In contrast to a deep drawing process of circular blanks, the process of drawing rectangular parts is more complex. The ambition is, to realize a forming process for deep drawing components without residual flange to save additional manufacturing steps. This is even more important for micro deep drawing since the size of the products complicate subsequent processing, such as trimming. In order to achieve parts with a net shape many blank shape optimizations were carried out in the macro range. As demonstrated in previous chapters, size effects due to miniaturization occur which prevent the unadjusted transfer of this optimizations to the micro deep drawing process. For instance, the friction coefficient increases significantly with decreasing process dimensions. The changes in friction and material behavior fundamentally affect the final geometry of the micro deep drawing parts. Hu [27] showed that by consideration of the actual friction coefficients and usage of the foil specific material properties in a size-dependent FEM-simulation, it was accordingly possible to optimize the blank geometry for rectangular deep drawn parts (see Fig. 5.8). Therewith the residual flange was minimized. In this connection, the material characteristics were identified as a more important factor than friction. Experimental investigations show good agreement with results from FEM-simulation and prove the flange-free drawability for rectangular parts down to a size of $1.5 \times 0.75 \text{ mm}^2$ for Al99.5. Further investigations confirm the transferability of these results to materials of E-Cu58 and 1.4301 [28]. It is also revealed that the blank holder force significantly affects the geometry of rectangular drawn micro parts.

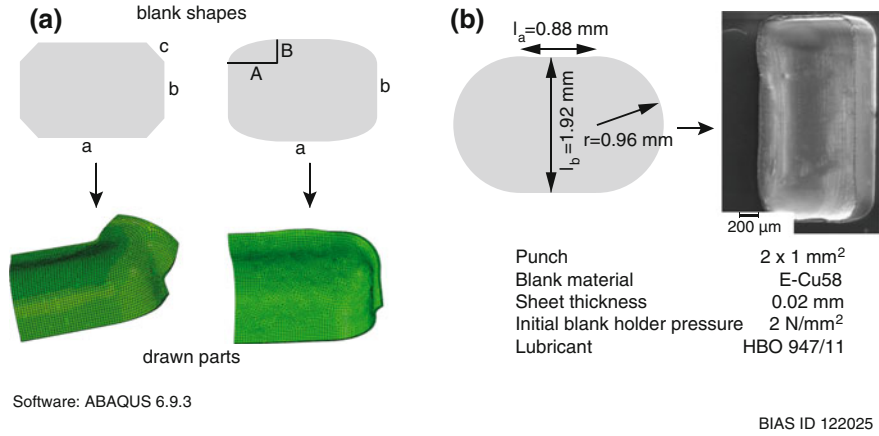


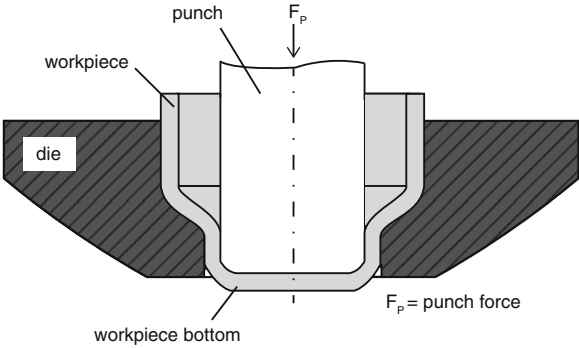
Fig. 5.8 **a** Optimization of blank shapes by FEM-simulation. **b** Rectangular deep drawn part with optimized blank shape

5.1.2 Multi-Stage Deep Drawing

In multi-stage deep drawing, a pre-drawn hollow shape is redrawn again, using several processing steps, to form parts with a smaller cross-section but with a larger drawing depth (see Fig. 5.9). This way it is possible to produce deep drawn parts with higher drawing ratios than realizable with a single step deep drawing process. The total drawing ratio is calculated as the product of the ratio of every single step. To achieve high values of the total drawing ratio an optimized incrementation, dependent on the process parameters, is necessary.

In Fig. 5.10 some examples of multi-stage micro deep drawing in industrial applications, contributed by Hubert Stueken GmbH & Co. KG, are presented. Exemplary components can be cathode cups for products such as LCD displays manufactured from $100 \mu\text{m}$ Ni-foils or electrodes for passenger cars brake lights, with an inner diameter of 1 mm and a total height of 9 mm . These examples show

Fig. 5.9 Schematic of multi-stage deep drawing



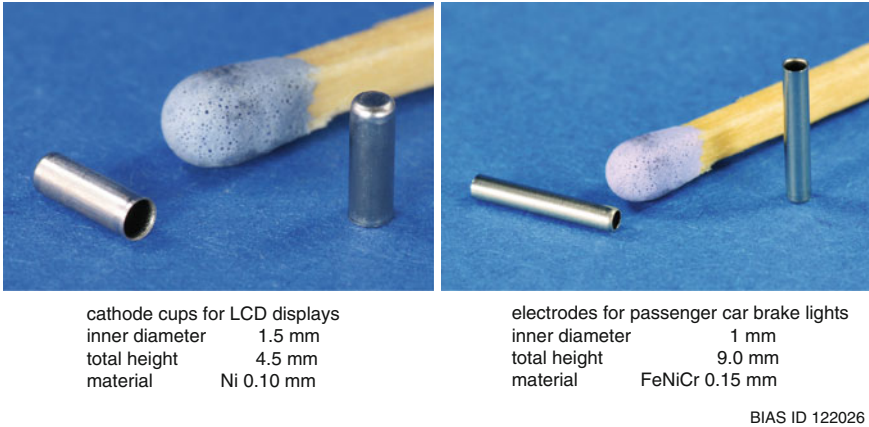


Fig. 5.10 Exemplary components from multi-stage micro deep drawing manufactured by Hubert Stueken GmbH & Co. KG

that the fabrication of high aspect ratios is technologically and economically feasible even in the micro range.

5.1.3 Scatter of the Punch Force

A further phenomenon that is frequently observed is an increase in scattering results with decreasing dimensions. This particularly concerns the determination of the material properties, such as the flow stress. In deep drawing similar discoveries are made for the scatter of the punch force. Reference [21] obtained deviations of about 20 % to the averaged punch force in micro deep drawing experiments using a 1 mm punch. In contrast, punch diameters of 5 and 10 mm only showed deviations of about 3 %. Reference [58] revealed a deviation of 37 % for deep drawing with a punch diameter of 1 mm and a drastic decrease in scattering punch force for a diameter of 5 mm down to deviations of 4 % (see Fig. 5.11).

Generally it is assumed that the increase in scattering with decreasing process dimensions is caused by the random orientation and size of each single grain leading to inhomogeneous material behavior and therefore to higher differences in the results [17]. But investigations by Justinger et al. showed that the scatter of the punch force cannot be explained by the orientation of the single grains. In [34], a model estimating the influence of the single grain orientations on the flow stress is presented. As a quantity for the variation in flow stress the Taylor-Factor was used. A statistical study analyzing the variance of the averaged Taylor-Factor, depending on the number of grains inside a unit volume, leads to the conclusion that significant scatter of the flow stress due to grain orientation is expected if less than 50 grains are inside the volume considered. The considered deformed volume

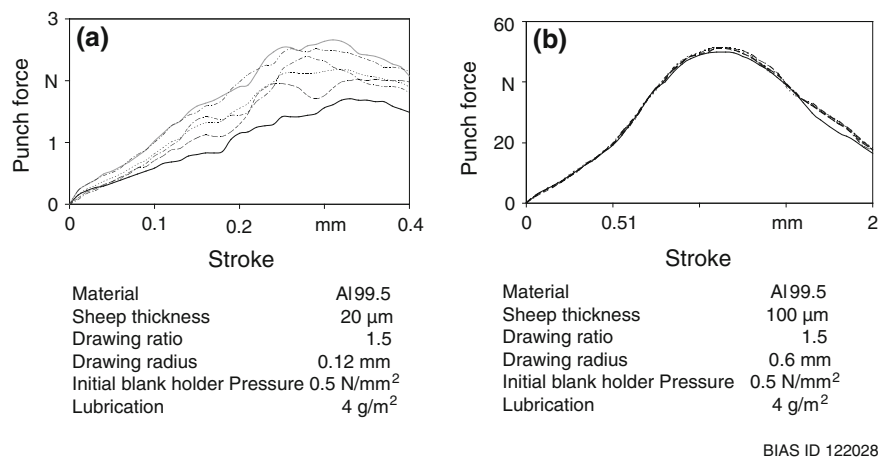


Fig. 5.11 **a** Deviation in punch force in deep drawing with 1 mm punch diameter. **b** Deviation in punch force in deep drawing with 5 mm punch diameter (both [58])

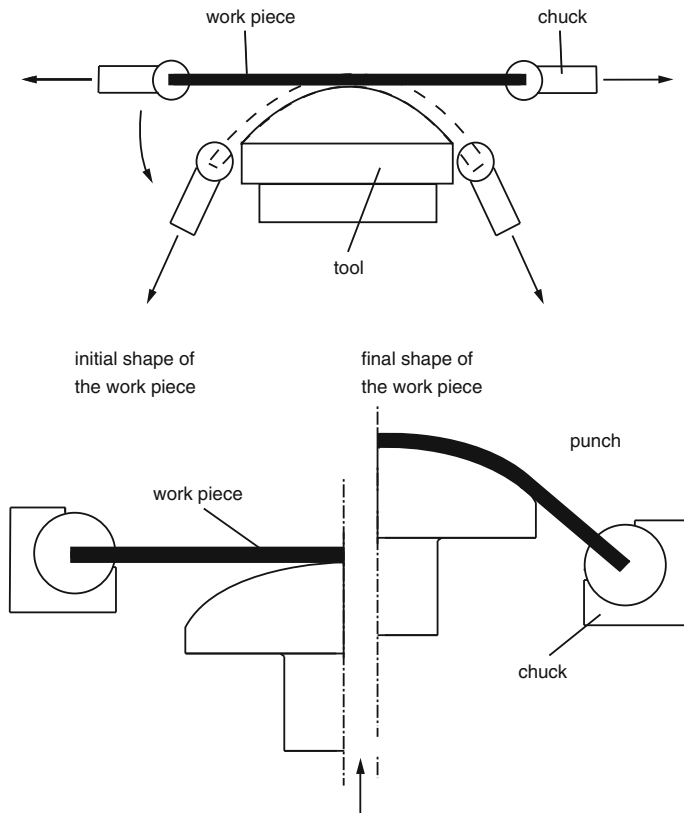
is for example the cross-section of a tensile specimen or a ring element of a circular blank for deep drawing multiplied by the sheet thickness. It is shown that such a ring volume generally contains at least several hundred grains in the relevant forming zone, even in micro forming. In addition, there are multiple forming zones occurring simultaneously in deep drawing. The punch force averages the stresses from all these different forming zones, and therefore it can be concluded that the scatter of the punch force is statistically not influenced by grain orientations and must have other sources. At the moment there is no generally valid explanation for the increased scatter in punch force with decreasing size. It cannot be excluded that the explanation might be found in the individual experimental conditions. For example, uncertainty in the punch force measurement would result in a higher relative measurement error for very small forces and might therefore be one possible explanation.

5.2 Stretch Drawing

Hanna Wielage

5.2.1 Mechanical Stretch Drawing

At mechanical stretch drawing a punch drives in a blank, similarly to mechanical deep drawing, and leads to the formation of a hollow body. Compared to mechanical deep drawing no flow out of the flange area is allowed. This condition can be the consequence of the tool geometry. Figure 5.12 shows two basic



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Fig. 5.12 Elementary stretch drawing (*left*) and tangential mechanical stretch drawing (*right*)

processes of the mechanical stretch drawing of sheets. In the case of elementary stretch drawing (Fig. 5.12, left) the tool is fixed at supports at the end of the workpiece and is formed by the moving punch. This leads to a strong friction constraint of the sliding movement between the punch and tool, and finally to a non-uniform strain distribution in the workpiece and early failure. These disadvantages can be avoided by tangential mechanical stretch drawing (Fig. 5.12, right), where the workpiece is pre-loaded and laid on the tool without any significant sliding movement. Thus the process is divided into tensile forming and bending.

Yamaguchi et al. used dies with different diameters in mechanical bulging tests in order to investigate the effect of thickness on the restoration of bulged specimens [64]. They performed experiments with sheets of different materials of thicknesses between 0.2 and 2 mm. Thereby it has been ascertained that the thinner the sheet metal, the greater is effect of buckling on the formation of wrinkles.

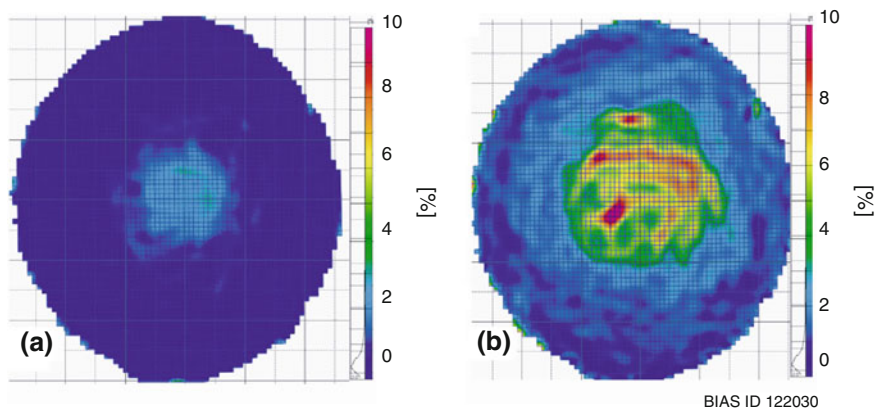
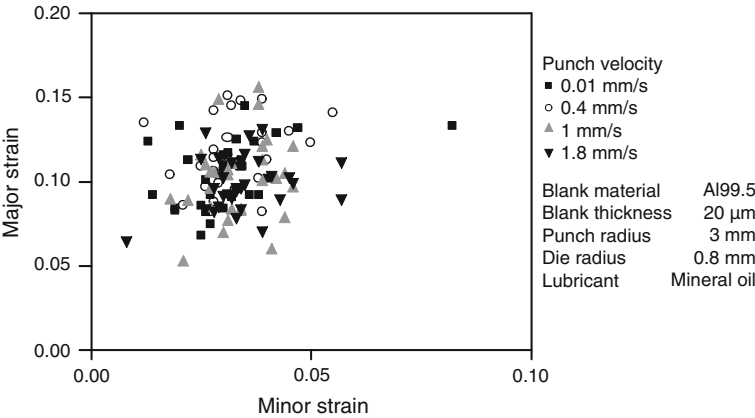


Fig. 5.13 Sample of stretch drawn part with punch velocity of 0.01 mm/s under different punch positions x : **a** $x = 0.39$ mm, **b** $x = 0.78$ mm, material: aluminum with thickness 20 μm

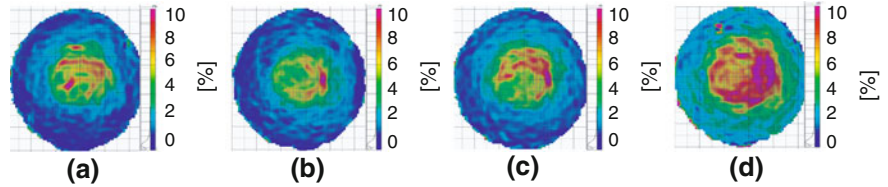
The deformation of mechanical stretch drawn parts with homogeneous microstructures is expected from FEM simulations to be rotational symmetric. Aluminum samples with a thickness of 20 μm show that stretch drawn parts do not deform with a rotational symmetric pattern as expected. The uneven strain distribution is also maintained for different drawing speeds [23]. The strain does not distribute uniformly, and mainly exists within the contact area between the punch and blank. Even within the contact area, the strain distribution is not uniform, especially at a relative deeper punch position, see Fig. 5.13b. As the punch moves into the die, the contact area as well as the strain of the specimen becomes larger. Some local strain maxima appear. These are starting points of the cracks that will appear later. No cracks show the same major and minor strains. Metallographic analysis has shown that samples with a thickness of 100 μm with a polycrystalline structure have a better forming behavior (according to the forming limit diagram) than samples with a thickness of 20 μm with a Tiffany structure [25]. The Tiffany structure amplifies the localization of strain and early failure (bottom fracture) occurs due to the excessive straining of single grains (Sect. 2.3). The forming limit diagram for Al99.5 with a thickness of 20 μm for 4 different punch velocities (from 0.01 to 1.8 mm/s) (Fig. 5.14) shows that there is no influence of punch velocity on the forming limit diagram.

With higher punch velocities the height of the stretch drawn parts increases [23]. In Fig. 5.15 the final frames of specimens before cracks under different punch velocities are displayed. For all punch velocities, the strain maxima are nearly the same while the strain distributions differ from each other. At the lowest punch velocity (Fig. 5.15a) the strain mainly exists within the contact area between the punch and blank. As the punch velocity increases, there is more and more strain outside this contact area. This means that more material is involved in forming when the punch velocity is increased, respectively, the strain rate. Therefore a larger height of parts can be obtained with increasing the punch velocity.



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Fig. 5.14 Forming limit diagram of Al99.5 with a thickness of 20 μ m for different punch velocities



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Fig. 5.15 Influence of punch velocity on the strain distribution of stretch drawn parts before cracking: **a** $v_p = 0.01$ mm/s, **b** $v_p = 0.4$ mm/s, **c** $v_p = 1$ mm/s, **d** $v_p = 1.8$ mm/s, material: aluminum with a thickness of 20 μ m

5.2.2 High Speed Stretch Forming by Shock Waves

The first verification of the possibilities to generate laser-induced shock waves in laser material processing was carried out by White at the beginning of the 1960s [61]. The first use of shock waves was in laser shock peening, what was in one row with sharpening, mechanical shot peening, water-jet hardening and hardening by cavitation bubbles, which also works with induced shock waves for the local hardening of the material. Basic investigations into material process technologies by laser-induced shock waves with the aim of building up residual stresses were conducted between 1968 and 1981 at the Battelle Institute in Columbus, Ohio, USA. Since 1986, numerous studies have been performed in France with the help of the automobile industry in the field of laser material processing by laser-induced shock waves in order to increase the lifetime of highly stressed metallic components by inducing residual stresses and thus developing an effective process [46].

Other areas of application of laser-induced shock waves include preloading of μm -thick components [10] or micro-perforating [4].

In forming technologies laser-induced shock waves were already used for research purposes. Gao et al. presented the new process of Micro Scale Dynamic Forming (μLDF) for the production of complex 3D-shapes in thin metal foils [16]. By using q-switched Nd:YAG-laser-induced shock waves, copper foils with thicknesses between 3 and 15 μm they formed cuboids with dimensions of 50 times 30 μm with a depth of 17 μm . Liu et al. produced micro channels in 10 μm thick copper foils with dimensions of 260 μm times 50 μm with one laser shock pulse. On 15 μm thick aluminum foils they only produced components with fractures and ablated surfaces [41]. Sagisaki used a femtosecond-Ti:Saphir-laser for the forming of 50 μm thick aluminum foils. He produced pyramids with a width of 30–75 μm and a depth of a few micrometers with 120 laser shock pulses. All these processes work with a sacrificial layer.

Shock waves used in laser material processing are usually initiated by Nd:glass- and Nd:YAG-lasers. Thus, the laser beam hits the workpiece surface and the laser pulse energy is absorbed by the material. Thereby ablation occurs as well as a metallic vapor cloud, which is ionized by the laser radiation [1]. In order to avoid damage to the surface by free ablation a sacrificial layer is used, which is completely ablated by the laser treatment. The coupling of laser radiation occurs by a transparent layer (glass or water) [44]. Both layers contribute to a 4- to 10-fold increase in shock wave pressure (depending on the laser pulse intensity) in comparison to laser plasma initiated in a vacuum [14]. The application of separated layers is especially necessary for the treatment of visible surfaces, and for the insertion of residual stresses. The undesirable effect of material removal with the initiation of shock waves by lasers can be avoided by a significant increase in the wavelength and a process-specific increase in laser pulse intensity [30]. Vollertsen et al. show this by means of a comparison of surfaces, treated by different types of laser radiation: In comparison to the use of an excimer-laser ($\lambda = 248 \text{ nm}$) with the use of a TEA- CO_2 -laser ($\lambda = 10.6 \mu\text{m}$) the ablation of aluminum can be avoided [51]. Also Bergmann and Hgel focused on the ablation behavior on surfaces by plasma induced by different laser systems. Thereby they detected that by using a TEA- CO_2 -laser (pulse duration: 120 ns, laser pulse energy: $<10 \text{ J}$, focus diameter: 1–10 mm) in air with laser pulse intensities $>10^8 \text{ W/cm}^2$ there was no change in surface quality on aluminum surfaces [5]. The inverse bremsstrahlung increases with the square of the wavelength, so that the longer wavelength of the CO_2 -laser is nearly completely absorbed by the plasma. This kind of shock wave formation has been known since the 1970s, e.g. Barchukov et al. [3]. Experiments on forming processes by CO_2 -laser induced shock waves are only known by Vollertsen et al. This forming process will be explained in detail in the following.

First of all, the initiation of plasma requires the existence of free electrons in the optical field of the short pulse laser. Thereby free electrons are generated out of the surface by thermo or field emission through laser treatment [30]. The number of free electrons depends on the focus area, laser pulse intensity and surface material

[43]. These free electrons absorb energy by inverse bremsstrahlung absorption and can produce further ions and electrons by impact processes until it comes to an optical breakdown and thus plasma formation [50]. In this case, the ignition point of the plasma is above the workpiece (~ 5 mm [2]). With a higher laser pulse intensity ($>10^7$ GW/cm²) the power of the shock wave ahead is large enough to heat up the environmental gas in that manner, so that further laser radiation is absorbed [laser-supported detonation wave (LSD-waves)] [29]. If the energy density of the laser pulse exceeds a certain threshold, the fast expansion of the plasma propagates a shock wave [45]. This shock wave moves spherically [60] and with it so does the heat, which is dissipated by the plasma [2]. If the shock wave initiated pressure lays over the flow stress of the material, the shock wave pressure leads to a plastic forming of the surface, e.g. [66]. As a threshold value for the ignition of the plasma Maher determined in 1973 with a Marx Bank laser, a laser pulse intensity of 1.2×10^7 W/cm² for aluminum is considered.

In Fig. 5.16, the principle of the forming process by laser-induced shock waves, in the following referred to as laser shock forming, is outlined for laser shock stretch forming. The principle experimental set-up for laser deep drawing is described in the following. A laser cut circular sheet metal is placed on a die. The blank holder is placed onto the blank with a defined blank holder force over the clamping force. In a next step, one or several short laser pulses of a TEA-CO₂-laser (wavelength 10.6 μ m, pulse length 100 ns, pulse energy 5.6 J) irradiate the specimen with the focus located on the blank surface. The formation of the shock wave takes place as already described previously for the high laser pulse intensity. This shock wave leads to a forming of the surface [52]. Depending on the material the surfaces show different influences of laser treatment. While with aluminum the surfaces show no influence on the surface, copper surfaces show a little discoloring. But even after 300 pulses no melting of the surface is observed. On austenitic steel, stronger discoloration is observed. Such discoloration results from N-Fe-bondings and oxides, which are built by the laser pulse treatment. In ambient air and argon atmospheres this discoloration can be reduced.

In Fig. 5.17 images of samples of stainless steel with thicknesses of 20 and 50 μ m and in Fig. 5.18 images of samples of copper and aluminum with a thickness of 50 μ m are shown. The samples are treated with similar parameters.

The sample of stainless steel with a 20 μ m thickness (Fig. 5.17a) shows minimal deformation of 100 μ m in cup height. The sample with 50 μ m thickness (Fig. 5.17b) shows no visible deformation after being subjected to 700 pulses. After 700 pulses, high damage to the material is noticeable, but still no deformation occurs. In contrast, the 50 μ m copper sheet can be formed, as well as can aluminum (Fig. 5.18a). With the same parameters, copper shows a 1.5 times smaller cup height than aluminum (Fig. 5.18b), which is due to the higher flow stresses of the material. However, high forming degrees up to the forming limit can also be reached with copper by the use of more pulses.

Since the yield stress of copper is comparatively low, the forming of copper is much easier to achieve. The higher ductility of copper in comparison to aluminum results in a smaller elongation ratio, but with a later crack formation than

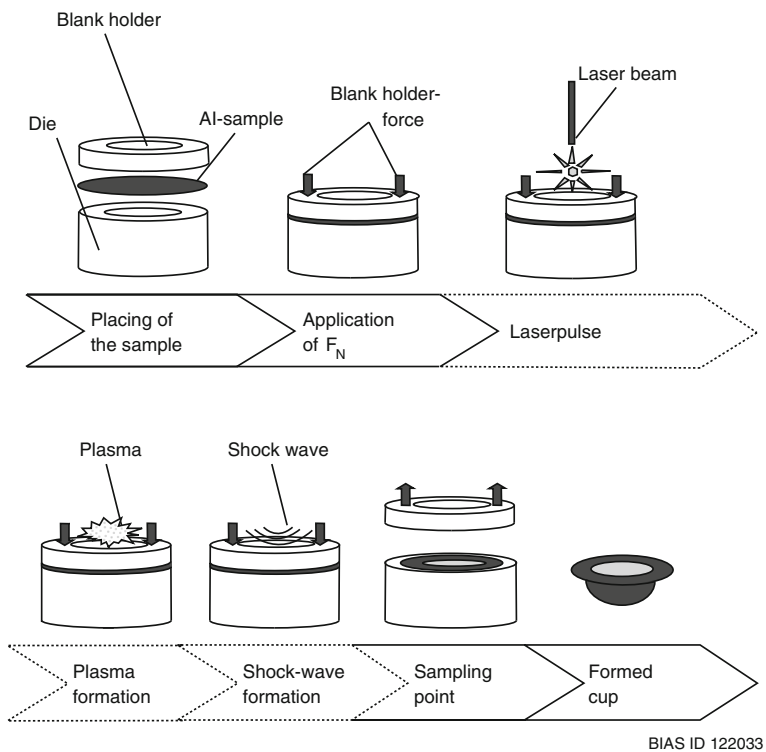


Fig. 5.16 Principle of the laser shock forming process

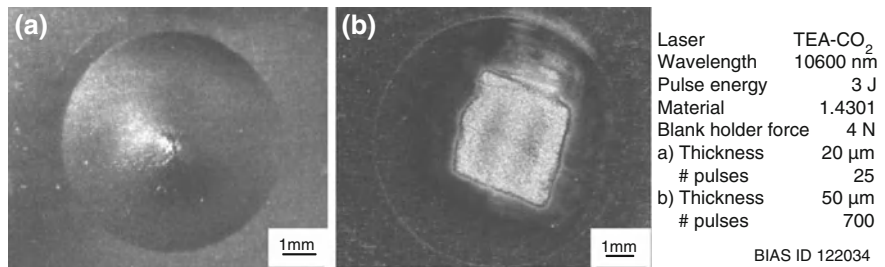


Fig. 5.17 Stainless steel samples with different numbers of pulses

aluminum. This shows the further potential for the use of copper in laser shock forming. Thus the application of laser shock forming with copper makes the process especially interesting for electronic industries.

The laser pulse energy that is spent in the forming process ranges from 3 to 6 J. In combination with the forming degree of the samples reached, the question of the efficiency-benefit-ratio arises. Assuming a full stretch-forming process the efficiency η is the ratio of forming energy E_F to laser pulse energy E_L :

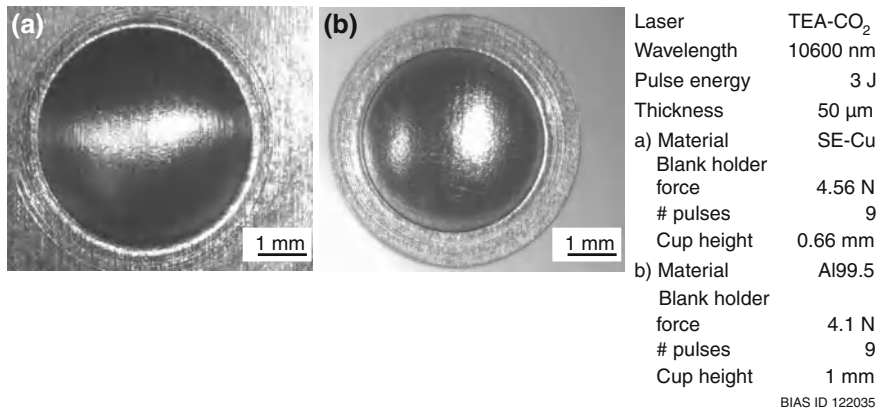


Fig. 5.18 Copper (a), and aluminum (b) samples with the same parameters

$$\eta = \frac{E_F}{E_L} \quad (5.1)$$

where the total deformation energy for full stretch-forming achieved is:

$$E_F = V \cdot k_f \cdot \varphi \quad (5.2)$$

The forming degree φ can be calculated from data of achieved stretch-formed workpieces, with s_0 and s_1 as the sheet thicknesses before and after pulse application, A_0 as the area of the sample before pulse application and A_1 as the generated surface of the formed cup after pulse application:

$$\varphi = \left| \ln \frac{s_1}{s_0} \right| = \left| \ln \frac{s_1}{s_0} \cdot \frac{s_0 \cdot A_0}{s_1 \cdot A_1} \right| = \left| \ln \frac{A_0}{A_1} \right| \quad (5.3)$$

With a die diameter of 4 mm with an achieved calotte height of 0.73 mm for a laser pulse energy of 3.5 J and 0.79 mm for a laser pulse energy of 5.7 J, the forming degrees are 0.12 and 0.14 respectively.

This method yields an average value for the true strain φ , which is useful to calculate the total deformation energy E_F . Assuming pure stretch drawing with forming degrees from formula (5.3) with a workpiece volume V of 1.26 mm³ and the flow stress k_f for aluminum ($s_0 = 100 \mu\text{m}$) of 27 MPa, the forming energy E_F for the laser pulse energy of 3.5 J is 4.1 mJ and for 5.7 J it is 4.8 mJ.

By the use of formula (5.1), efficiency can be calculated as 11.6×10^{-4} for a laser pulse energy of 3.5 J and to 8.4×10^{-4} for a laser pulse energy of 5.7 J. Taking into account the achieved drawing depth after one laser shock pulse, it is obvious that an increase in laser pulse energy leads to a decrease in the efficiency of stretch drawing. The influence of the material on efficiency will be clear when the efficiency for copper is determined. Pure stretch drawing efficiencies for copper with a thickness of 50 μm of 6.2×10^{-4} for a laser pulse energy of 3.5 J (with an

achieved conical workpiece height of 0.73 mm), and 4.7×10^{-4} for a laser pulse energy of 5.7 J (with an achieved conical workpiece height of 0.81 mm), could be determined.

It can be seen that the decrease in efficiency with increasing laser pulse energy for aluminum and copper is nearly the same. The efficiency is 1.3 times smaller than the efficiency of aluminum. On the one hand this can be explained by the increase in energy, which is necessary in order to hit the electrons out of the surface, and on the other hand a larger share is reflected instead of absorbed in the case of copper than in the case of aluminum. The latter leads to a higher energy loss in copper compared to aluminum, which in benefit turns to a lower energy loss effectively for copper [62].

Significant characteristic values of the laser shock forming process are the workpiece velocity and the strain rate. With a maximum laser pulse energy of 5.5 J and a workpiece velocity of 40 m/s, strain rates of $3 \times 10^3 \text{ s}^{-1}$ can be achieved. A further important characteristic value is the shock wave pressure. Here, pressure maxima of 8–20 MPa can be achieved [62, p. 105]. By the use of a conical blank holder, which limits the spreading of the shock wave, the pressure can be increased, since there is a dependency of the reachable workpiece shape and the blank holder shape. The reason for the dependency mainly lies in the dependency of the acceleration to the workpiece center. With longer pressure contact time material is taken from the sides and a spherical shape is built. Thus next to the pressure maximum, further parameters have to be considered for a maximum forming depth. Due to the constant pressure for defocusing above 0 mm, a correction of the focal position

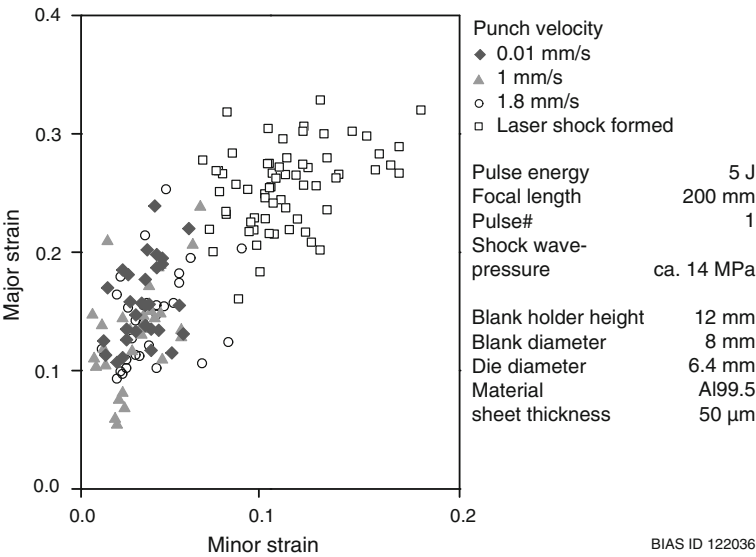


Fig. 5.19 Forming limit diagram for laser shock forming in comparison to mechanical stretch drawing with different punch velocities for Al99.5, $s_0 = 50 \text{ μm}$

Table 5.1 Process parameter and forming behavior characteristics in laser shock forming [62]

	Workpiece velocity for bending	Strain rate for bending	Pressure raise time	Forming behavior
Laser shock forming	40 m/s	$10^2\text{--}10^3\text{ s}^{-1}$	4 μs	Higher max. strains achievable than by quasi-static forming

during stretch drawing with multiple pulses is not necessary. Also, the adjustment of the focal position does not play an essential part.

A further process property applies to the formability behavior. In [62] it could be shown that in comparison to mechanical stretch drawing with different punch velocities, higher minor and major strains could be achieved by laser shock forming (Fig. 5.19). The main process parameters of the high speed forming process of laser shock forming are displayed in Table 5.1.

5.3 Bending

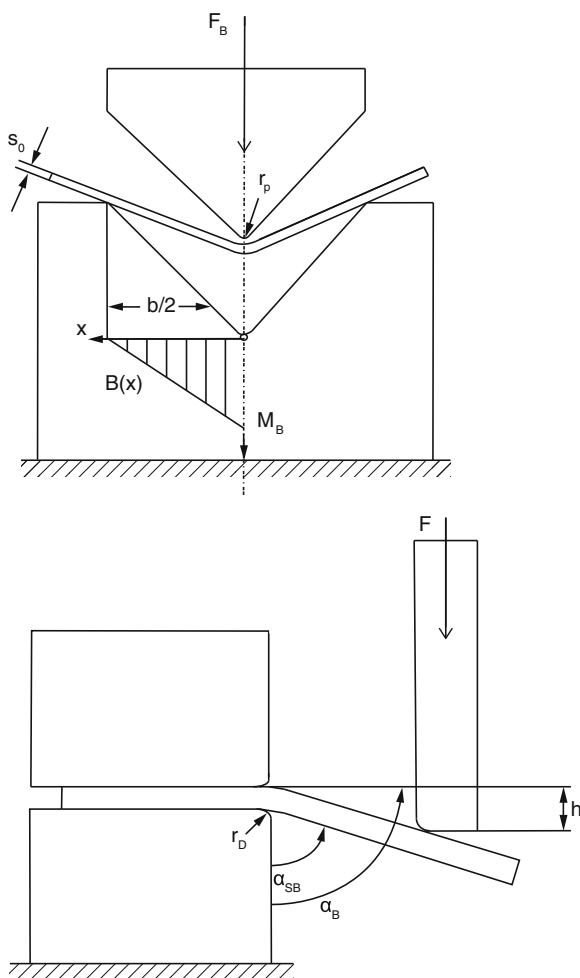
Hanna Wielage

The process of bending is often used in the production of micro system technologies (MST) or micro electro-mechanical systems (MEMS), i.e. for clamps or connectors, Fig. 5.20.

Fig. 5.20 Micro bending element (spring element from Fa. Harting GmbH & Co, Germany)



Fig. 5.21 Geometry, forces and moments of air bending (*upper*) and free bending (*lower*)



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Forming by bending is defined as the forming of a solid body, where angular or ring-shaped workpieces are produced. In this case the state of plasticity is mainly induced by a bending load. The most important bending processes are air bending and free bending (Fig. 5.21). Air bending can be either produced by a rigid clamping of one of the bending legs or by the bending of a sheet over a mid-point loading, which is placed on two supports. If the workpiece is completely rigid and clamped, a bended sheet occurs, which has an even leg and a material-dependent curved leg. Free bending starts with air bending. It relates to free bending as far as the workpiece has more contact with the tool than the 3 initial contact points. Stresses under loading are in equilibrium with the external forces, whereas removing external forces results in spring back, and thus in a change in stress

condition. This change leads to elastic spring back, which means a reduction of the bending angle after load removal compared to the bending angle under loading. The order of this spring back can be specified by the spring back ratio K , which relates the ratio of bending angle after load removal α_R to the bending angle under load α :

$$K = \frac{\alpha_R}{\alpha} \quad (5.4)$$

The spring back ratio greatly depends on the material behavior and the ratio of bending radius to sheet thickness. Variations in sheet thickness and material properties lead to a remaining variation in the bending angle after load removal.

An estimation of the elastic spring back angle can be carried out with the material data based on 0.2 % yield strength $\sigma_{p0.2}$ and the E-modulus, the tool geometry data based on the die radius r_D and bending angle α_B as well as the material thickness s_0 using the following formula (see [55]):

$$\alpha_{SB} = 3 \cdot \frac{\sigma_{p0.2}}{E} \cdot \frac{r_D}{s_0} \cdot \alpha_B \quad (5.5)$$

The acting bending force F_B , marking the beginning of the plastic flow at the outer layers, can be calculated by theoretical considerations. In air bending the maximum bending moment M_B acting during the purely elastic deformation at the center of the die can be approximately determined by:

$$M_B = \frac{1}{4} \cdot F_B \cdot b \quad (5.6)$$

where F_B is the bending force and b the initial die width. The corresponding stresses σ_x in the outer fibers at the upper and lower sides of a workpiece are:

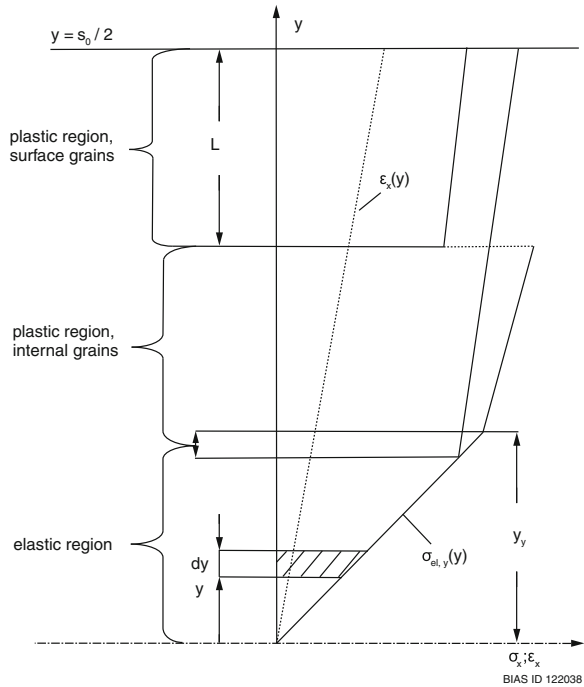
$$\sigma_x = \pm \frac{M_B \cdot s_0}{2 \cdot I_Z} \quad (5.7)$$

where I_Z is the inertia moment around the z-axis. With formulas (5.6) and (5.7), the acting bending force $F_{B,pl}$ marking the beginning of the plastic flow at the outer layers can be calculated with $\sigma_x = k_f$:

$$F_{B,pl} = \frac{8 \cdot k_f \cdot l_z}{b \cdot s_0} \quad (5.8)$$

Although these formulas enable an approximation of the acting bending force, approaches for the precise prediction of the acting bending force marking the beginning of plastic flow have not been successful [35]. With decreasing material thickness, the difference between the simulation and experimental results grows [8]. It is impossible to develop a universally applicable FEM model, although significant work has subsequently been undertaken on this subject [17]. These aberrations in the prediction of the bending force were explained by the different behaviors within the surface zones and the core zone, which is discussed in detail

Fig. 5.22 Stress-strain distribution during bending considering the deviating flow behavior of the surface and internal grains, according to [35]



in Sect. 3.3.4. Thus for a more specific calculation it is necessary to consider the different flow stresses of the grains on the surface and the inner grains in the calculation, see formula 3.23.

Figure 5.22 shows schematically the stress distribution during bending in one half of the workpiece’s cross-section for both cases [35]. Table 5.2 gives an overview of the acting bending forces F_B normalized over the effective width b for thin foils for different materials.

Table 5.2 Acting bending forces normalized on the effective width F_B/b in micro range for brass, copper and aluminum foils

Bending Material	Normalized bending forces F_B/b		
	Air bending Brass foil C2680 (annealed) [54]	Free bending Al99.5 foil (coarse grain) [9]	Free bending SE-Cu58 foil (coarse grain) [9]
s_0			
25 μm	–	0.4 N/mm	1.48 N/mm
40 μm	0.04 N/mm	–	–
50 μm	–	0.50 N/mm	1.25 N/mm
60 μm	0.10 N/mm	–	–
80 μm	0.20 N/mm	–	–
100 μm	0.24 N/mm	1.10 N/mm	2.98 N/mm
200 μm	–	1.96 N/mm	6.08 N/mm

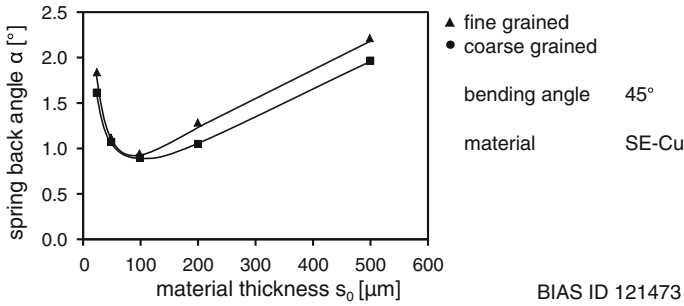


Fig. 5.23 Dependence of spring back angle on scaling factor with a bending angle of 45° (data with permission from [8])

In addition to difficulties in the precise prediction of the bending force, the prediction of the resulting bending angle, and thus elastic spring back angle, is one of the great challenges. Two main effects can be observed. Firstly, the order of the spring back angle behaves differently than in the macro range: the thinner the material the larger the spring back of aluminum. For copper foils the spring back angle increases with material thickness (Fig. 5.23) [9]. Two size effects related to bending processes are responsible. The first is caused by an increasing share of surface grains in the overall volume within thinner foils. The resulting effect is a decrease in material strength, because of the shape sum effect (Sect. 2.1.3), and thus a decrease in spring back angle. The second size effect is caused by large strain gradients appearing when the foil thickness is decreased, which results in a larger spring back. For aluminum the effect of strain gradient is predominant and leads to a constant increase in spring back with decreasing foil thickness. For copper, the effect of the share of surface grains dominates down to a scaling factor of 0.2. For thinner foils the influence of strain gradient is stronger, leading to an increase in spring back angle and bending moment (Fig. 5.23) [9].

The second effect in bending processes is that with decreasing material thickness the scatter of the spring back angle and thus of the scatter of the bending angle increases. Liu et al. give as a reason for this behavior the elastic anisotropy of the surface grain due to differences in grain orientation. The order of spring back strongly depends on grain orientation. In [42] they determined the spring back angle of foils with a thickness of $100\ \mu\text{m}$ and a grain size of $50\ \mu\text{m}$ with different grain orientations. The results show that the spring back angle of a foil with $[1\ 0\ 0]$ grain orientation is 25 % larger than that of a foil with $[1\ 1\ 1]$ grain orientation. Furthermore, they could observe an increasing scatter of spring back angle with decreasing foil thickness. The low amount of grains in combination with a strong difference in the orientation of the different grains results in an increase in scattering with a decrease in foil thickness.

With much higher workpiece velocities a further aspect can be observed in spring back behavior. High speed workpiece velocities can be realized by laser shock forming (Sect. 5.2.2) with which bending is also possible (Fig. 5.24). After

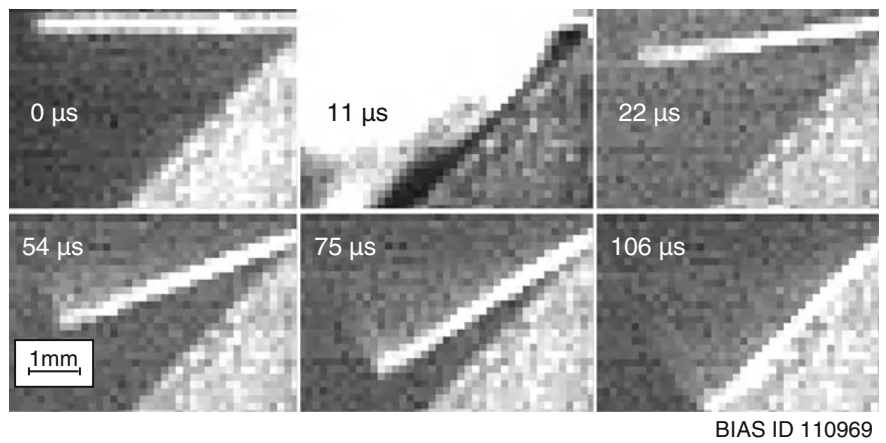


Fig. 5.24 High speed camera records of laser shock bending of workpiece hit with one TEA-CO₂-laser pulse, pulse energy: 5.6 J, material: Al99.5, $s_0 = 50\text{ }\mu\text{m}$

the workpiece hits the bending tool, spring back occurs. Thus a re-bending angle $\hat{\alpha}_R$ is caused, which is composed of the elastic share of the bending, the spring back with the spring back angle α_{SB} , and the share of rebound with the rebound angle α_{RB} . This occurs through the impact of the workpiece on the tool at high speed. The high speed bending of aluminum workpieces of different sizes show that with a thickness of $50\text{ }\mu\text{m}$ higher workpiece velocities can be reached than with a workpiece thickness of $20\text{ }\mu\text{m}$ (Table 5.3). In combination with the known dependency of spring back angle with the ratio of bending radius to workpiece thickness [formula (5.5)] an increase in spring back angle with increasing foil thickness would be expected, providing the same die radius is used. The evaluation of laser shock bending experiments (Fig. 5.25) results in contradictory behavior: despite a lower workpiece velocity, larger re-bending angles are reached for a workpiece thickness of $50\text{ }\mu\text{m}$ than for a thickness of $20\text{ }\mu\text{m}$ (Table 5.3) [62].

The reason for this behavior deviating from the theoretical observation is given in [62]. Experimentally-based and theoretically-based calculations of the occurring rebound angle are conducted. The theoretically-based calculation is done by consideration of the acting energies of the workpiece after hitting the bending tool. Thereby the energy for plastic forming of the workpiece, which is required for the generation of the rebound angle after impacting the tool, correlates with the kinetic

Table 5.3 Determined re-bending angles $\hat{\alpha}_R$ and return bending velocities v_R for thicknesses of 20 and $50\text{ }\mu\text{m}$ by laser shock bending (σ —standard deviation); for experimental conditions see Fig. 5.25

$s_0\text{ (}\mu\text{m)}$	$\hat{\alpha}_R$	v_R
20	$6.8^\circ\text{ (}\sigma = 0.9)$	$2\text{ m/s (}\sigma = 0.6)$
50	$12.1^\circ\text{ (}\sigma = 0.8)$	$4.9\text{ m/s (}\sigma = 0.8)$

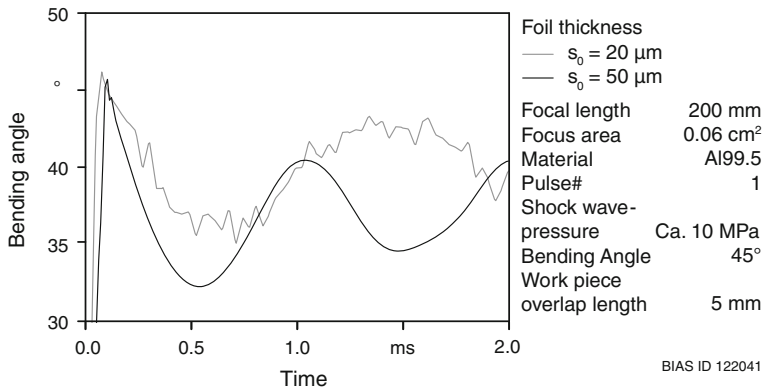


Fig. 5.25 Bending angle achieved by laser shock bending dependent on foil thickness for Al99.5 workpieces with thicknesses of 20 and 50 μm

energy, which acts after the impact and both are equated. The different calculations demonstrate that the rebound of the workpiece causes this behavior. Thus the rebound during the high speed forming of thin foils has to be observed.

5.4 Flange Forming

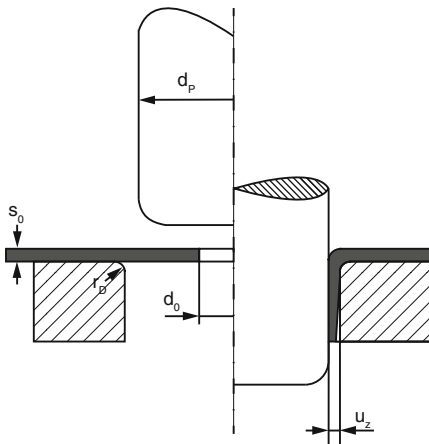
Heiko Brüning

The process of flange forming belongs to tension compression forming manufacturing methods. The workpiece, e.g. sheet metal foil, is placed between a blank holder and a perforated die, thus allowing penetration by the punch so that a closed flange is formed. In contrast with deep-drawing, plastic material flow beneath the blank holder is not desired. The radial deformation is rather small so that the length of the bending leg is constant. The forming operation is completed as soon as no sheet metal is left underneath the bottom punch surface.

Macro range flange formed geometries are often used as preforms for threads. This is a relatively cheap manufacturing method for large batch sizes because special tools are available that generate the flange and thread in one operation. Other applications of flange forming are in the increasing torsional stiffness of blanks as well as welding on tubes. Furthermore, flanges are used in order to give bolts or axles guidance and contact faces in sheet metal. These applications can be easily transferred into micro range, if reasonable.

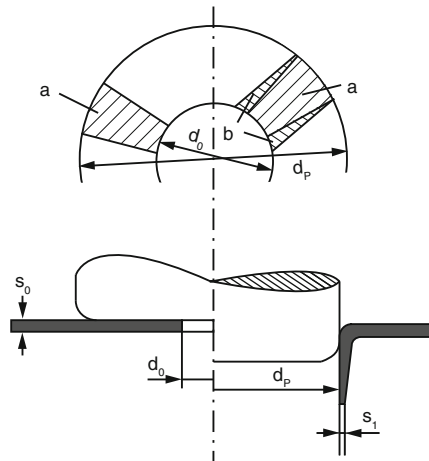
As stated above, the main difference between deep-drawing and flange forming from the applicator's point of view is the fact that the influence of the outer shape of the blank is almost negligible for flange forming, with the consequence that the blank holder force does not need to be adjusted as precisely and the outer geometry of the blank is not influenced by the manufacturing operation. A plastic material

principle



s_0 : initial blank thickness
 s_1 : minimum blank thickness
 d_0 : diameter of pre-hole
 d_p : punch diameter

forming a flange



a : blank material to be bent
 b : additional material due to reduction of thickness
 u_z : drawn clearance
 r_d : die radius

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Fig. 5.26 Principle of flange forming with pre-hole

flow is mainly induced by tangential stresses within the flange section, so that the material for forming the flange is provided by the thickness of the blank (see Fig. 5.26).

Flange forming is mainly carried out with sheet metal as wrought material, but tubes can be used as well [20]. In this paragraph tubes are not addressed, because in micro range flange forming the use of tubes has not yet been reported.

For both the micro- and macro range it is reasonable to divide the manufacturing method of flange forming into three different groups based on the tools needed for the process [56]:

- flange forming without pre-hole
- flange forming with pre-hole
- flange forming with pre-hole and backup punch.

For flange forming without pre-hole, the blank is fixed between the blank holder and die. The cone-shaped punch is driven through the blank and thus leads to it cracking. The location of the crack line is generally not determined so that the process tolerances have to be large. Due to the fact that the rim of the flange is formed by the crack line, flanges usually have a poor quality. Therefore flange forming without a pre-hole is rarely employed.

The quality of the rim can be improved by flange forming with a pre-hole. In this case holes are cut into the blank before the forming process takes place. It is obvious that the conditions of the hole borders have a strong influence on the result of the flange forming process [63]. The pre-hole is expanded due to the coaxial penetration of the punch, while the thickness of the blank is reduced. Flange forming with pre-hole is mainly affected by the sheet metal material and the quality of the pre-hole, but punch geometry as well as the scaled pre-hole diameter d_0/s_0 also have an influence.

The forming limit is reached as soon as the plastic material flow either leads to radial cracks in the flange or peripheral cracks at the die rounding. Peripherally oriented cracks appear if punches with a flat bottom surface and a very narrow drawn clearance are used [13]. In general, radial cracks are reported more often. Improving the material flow can be achieved by flange forming with a pre-hole in combination with a backup punch. The backup punch enables the possibility to increase the axial stress at the border of the pre-hole thus superposing tangential stresses, and therefore higher flanges can be formed because flanges generally tend to fail due to tangential stresses as metal materials tolerate lower tangential than axial stresses (Fig. 5.27).

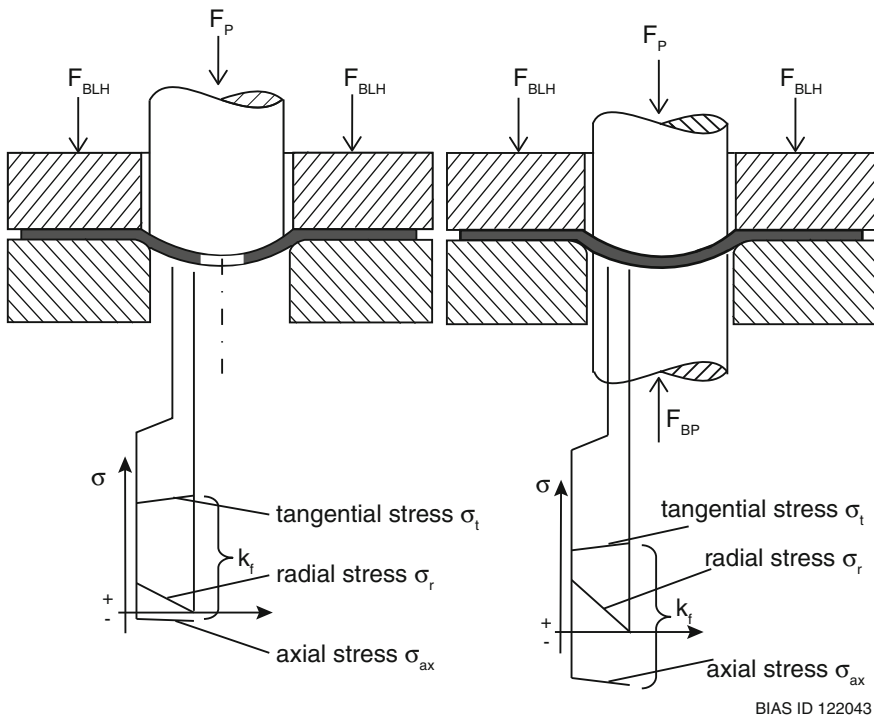


Fig. 5.27 Flange forming without and with backup punch and resulting stress states

The fact that the pressure between the punch and backup punch by a given force is dependent on the interfacial contact area, the bottom surface of the punch is assumed to be flat with a small-sized flanging radius to enable the maximum contact area.

The punch force needed for a complete forming operation is not only dependent on the sheet metal material and the expected flaring ratio, but also on the punch geometry [40]. Punches with a flat bottom surface generally require a higher punch force than punches with a truncated cone surface, hemispherical surface or tractrix surface. This is due to the fact that, considering at the tractrix shaped surface, the bending leg is always as large as possible thus decreasing the punch forces required [37]. Nevertheless, the advantages of a punch with a flat surface are that the displacement of the punch is the lowest and the manufacturing process is relatively easy. The larger the drawn clearance, the lower the influence of the punch geometry.

It is reasonable to differentiate between narrow flanges and wide flanges, depending on the ratio of the (inner flange diameter d_i)/(initial sheet thickness s_0). Lange [39] defines narrow flanges as those with $d_i < 5s_0$ and wide flanges with $d_i > 5s_0$.

Investigations by Schlagau [53] show that the flange height h of flanges in the macro range can be calculated as follows (see Fig. 5.26 for notation):

$$h = \frac{dp + 2u_z - d_0}{2} + 0.4r_D + 0.2s_0 \quad (5.9)$$

For ratios other than those given, flange heights can be calculated as shown by Romanowski [49].

For flange forming with a pre-hole, the flaring ratio is defined as d/d_0 . The maximum achievable flaring ratio is, as stated above, dependent on the sheet material and the quality of the pre-hole. Experiments in the flange forming with pre-holes have been carried out using sheet metal foil of austenitic steel 1.4301 with thickness $s_0 = 25 \mu\text{m}$ and Copper E-Cu58 with $s_0 = 20 \mu\text{m}$ and $s_0 = 10 \mu\text{m}$. For sheet metal foils with thickness $20 \mu\text{m} \leq s \leq 25 \mu\text{m}$, the punch diameter $d_p = 1.00 \text{ mm}$ and punch flange radius as well as the die diameter and die radius $r_D = 100 \mu\text{m}$ were kept constant, while the pre-hole diameter was gradually reduced thus increasing the flaring ratio starting from 1.1 and ending at 3.4. The pre-hole was generated by laser sublimation cutting using a Nd-YAG Laser. The punch velocity was 1.0 mm/s . No lubrication was applied. The same procedure was conducted for sheet metal foil of $s_0 = 10 \mu\text{m}$, but the punch diameter as well as the die radius were reduced to $d_p = 0.50 \text{ mm}$ and $r_D = 90 \mu\text{m}$ so that the ratio s_0/d_p was kept almost constant.

The maximum flaring ratio for 1.4301 $s_0 = 25 \mu\text{m}$ and E-Cu58 $s_0 = 20 \mu\text{m}$ achieved with the setup used was 1.6. Flange forming with E-Cu58 $s_0 = 10 \mu\text{m}$ was possible up to 1.3. For larger flaring ratios, cracks come into existence starting at the rim of the pre-hole. In the macro range these values are generally larger, up to 3.0 [13]. It is most likely that the experimentally determined relatively low

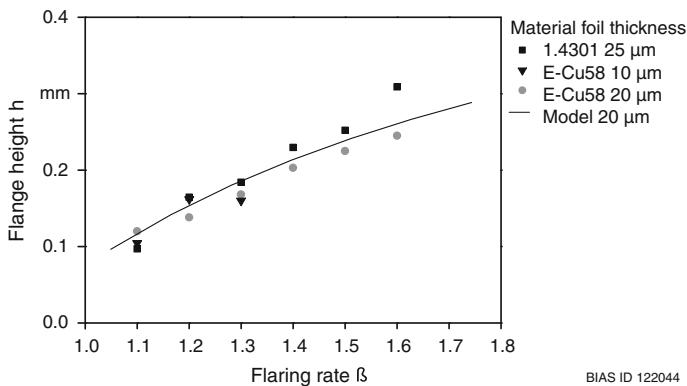


Fig. 5.28 Flange heights in dependence of flaring ratio; theoretically and experimentally determined values

flaring ratios are a consequence of the pre-hole quality, which is affected by oxidation due to the laser ablation process, which leads to the conclusion that the cutting method plays an important role in flange forming.

The resulting flange heights have been optically measured by a confocal microscope, with the results shown in Fig. 5.28. Compared to the theoretical model, which is based on formula 5.9 and approved for the macro range, the measured flange heights are generally smaller. Nevertheless, the experiment shows that the theoretical model is also applicable in micro ranges, as deviations between the measured values from processed flanges and predicted values are rather small.

Figure 5.29 shows a formed flange of 1.4301 with sheet thickness $s_0 = 25 \mu\text{m}$ with a flaring ratio of 1.6, as well as a cross-sectional view. It can be seen that the



Fig. 5.29 Material: 1.4301, size25, flaring ratio 1.6, $d_p = 1.0 \text{ mm}$. *Left* Formed flange. Rim is uneven due to oxidation effects of pre-hole generation by sublimation cutting. *Right* cross-sectional view, V2A-etchant

flange does not have any defects and that the flange height is almost constant at the circumference while the initial sheet thickness is reduced.

5.5 Piercing and Blanking

Gerrit Behrens

5.5.1 Piercing and Blanking Basics

Piercing and blanking are cutting processes by the application of a shearing force. When the shear stress in the material exceeds the ultimate shear strength, the material fails locally in the cutting zone and is subsequently separated. Shear cutting is not a metal forming process, but the cutting is always connected with a plastic deformation before the material finally fails. Piercing and blanking are cutting processes with a closed cutting path. They require a metal sheet or foil, a punch, a die and in most cases a blankholder (see Fig. 5.30). Both processes are very similar. In blanking the punched-out material is the desired workpiece and the remaining sheet is scrap, while in piercing, also known as punching, this remaining material with a desired inner profile is the workpiece and the removed material is discarded. Various shapes of the cutouts can be realized, reaching from simple geometric shapes such as circles or rectangles to combinations thereof, and more complex structures. There is also the opportunity of using a large number of punches and corresponding dies simultaneously to realize perforations in one single operation.

The actual shear cutting process is quite complex, involving shares of upsetting, bending, shearing and crack initiation and propagation, and it also depends on the material properties as well as the tool geometries. Therefore, these properties also determine the shape of the sheared edge. A characteristic cross-sectional profile of

Fig. 5.30 Schematic of the blanking and piercing process; according to [15] (with kind permission of Springer Science + Business Media)

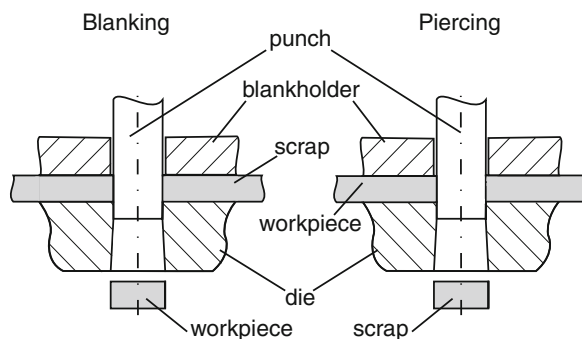
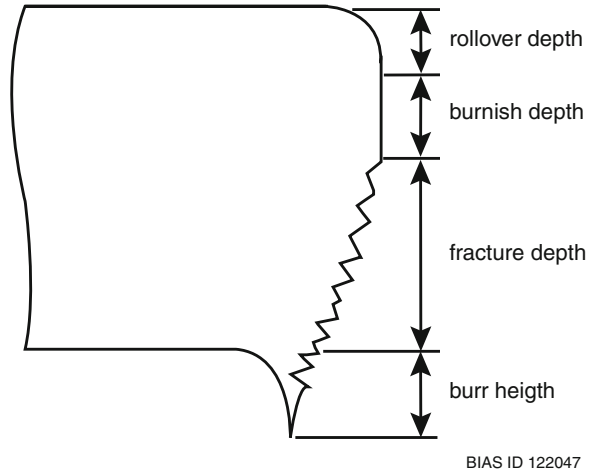


Fig. 5.31 Distinctive features of the sheared edge in piercing and blanking



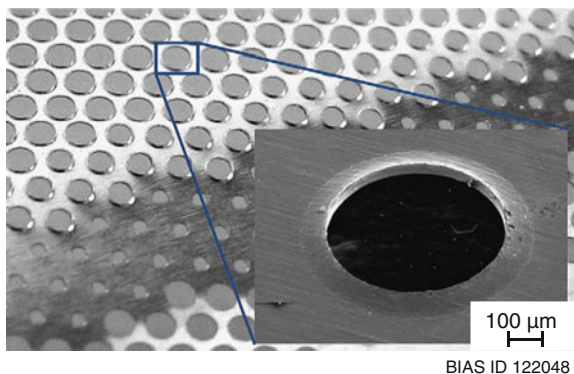
a sheared surface usually consists of rollover, burnish, fracture and burr. Rollover occurs when the punch contacts the sheet metal and draws the material slightly into the clearance. Further penetration of the punch leads to shearing and the material becomes locked and burnished in the punch and die clearance. When the formability in the shearing zone is exhausted, crack initiation occurs, which is followed by the final fracture of the material, and leads to a typical fracture plane as shown in Fig. 5.31. The distinctive features of the sheared edge are indicators of the process quality and may be important if subsequent forming processes are planned.

Piercing and blanking processes have large industrial relevance in the manufacture of micro parts [17]. Potential fields of application are leadframe structures, connecting the die of a microchip with a circuit board in the electronics industry, and injection nozzles for combustion engines or rinsing foils. A further industrially-relevant application for the use of micro holes is the outer shaving foil of electric shavers (see Fig. 5.32) [38] or the production of inkjet printhead nozzles [31]. But the miniaturization of the process is not applicable without any restrictions.

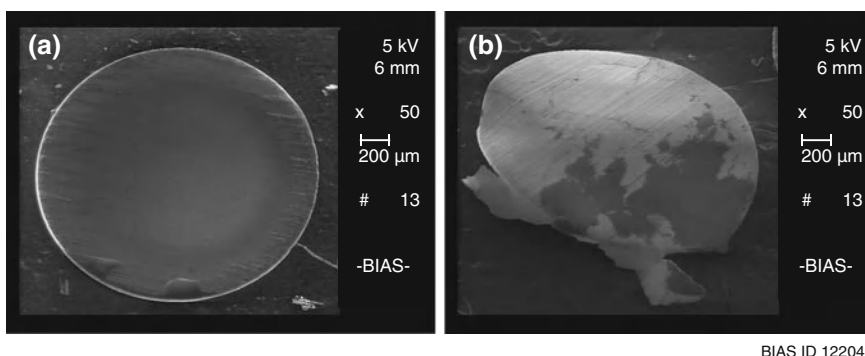
5.5.2 Tools for Piercing and Blanking

In piercing and blanking in the micro range there are two main technical obstacles that complicate the miniaturization. One is the fabrication of tools with a high geometrical accuracy and the other is the accurate alignment of the tools [65], since both the clearance between punch and die, as well as the alignment of the tools, are very important parameters to determine the quality of the piercing or blanking process. Under standard blanking process conditions the ratio of die clearance to workpiece thickness is about 5 %. This means the required die

Fig. 5.32 Manufactured prototype of an outer shaving foil with SEM picture of a detailed conically formed single hole [38] (reprinted with permission from Elsevier)



clearance for a 20 μm thick foil is 1 μm . Therefore, the alignment between a punch and die hole should be within 1 μm , and the straightness error in the punch motion must be less than 1 μm during a punch stroke of several millimeters [32]. This is quite a challenge for adjusting the die set. The manufacturing of such die sets requires advanced accuracy and time. An example for the susceptibility of a micro blanking process is shown in Fig. 5.33. Here, the blanking tool set was used in a follow-on tool, combining a blanking and a following a deep drawing process. The tool set for the blanking operation was manufactured as clearance fit (H7/g4) according to ISO 286 with a diameter of 1.7 mm. This results in a tolerated die clearance of at least 1 μm , in the case of truly accurate manufacturing and an ideal alignment, and at most 15 μm for suboptimal manufacturing and adjustment of the die set. The same tool set was used for the blanking of different foil materials (Al99.5, E-Cu58, 1.4301) and foil thicknesses (20, 25, 50 μm). For all tested materials with a thickness of 50 μm , the produced blanks exhibited high quality cutting shapes and cutting edges (see Fig. 5.33a). In contrast, a reduction in the foil thickness by a factor of 2 or 2.5 always led to defective blanks (Fig. 5.33b). This demonstrates the sensitivity of a piercing or blanking process to changes in



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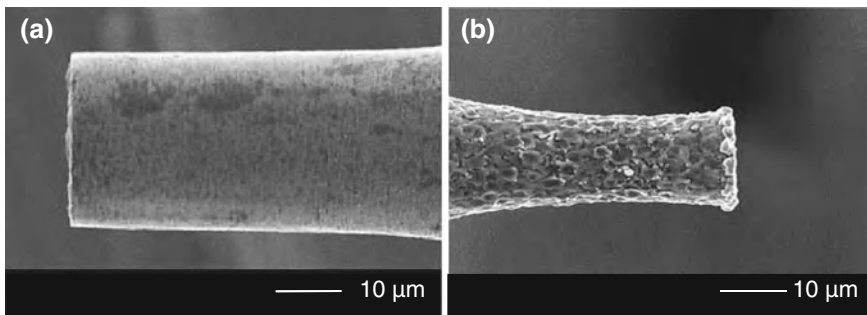
Fig. 5.33 Cu-blanks with different thicknesses produced by blanking using the same tool set. **a** Foil thickness 50 μm . **b** Foil thickness 20 μm

material thickness, and also illustrates the necessity of high geometrical accuracy in the manufacture of the tool set.

So there is an increase in demand for innovative and very precise manufacturing technologies to realize these challenges. Special machines with raised precision as well as improved existing manufacturing processes, or even new ones, for the production of the tools are necessary to guarantee a high quality piercing or blanking process in micro scale applications. Therefore several tooling concepts were developed, and studies on micro hole piercing were investigated.

Conventional methods of tool manufacturing can, with restrictions, also be used for micro forming tools. However besides manufacturing accuracy problems, the tool's availability, for example for micro drilling, is very limited and therefore also restricts the fabrication of dies for piercing and blanking [7]. Reference [32] reports on small punches made by ultra-precision micro grinding, achieving punch diameters as small as 25 μm (see Fig. 5.34a). However, for producing noncircular tools other processes have to be used. Electrical discharge machining processes (EDM) enable the manufacture of even smaller structures without limitations on circular geometries, as shown in [7]. Wire-EDM offers the opportunity of generating piercing and blanking dies down to a diameter of 10 μm [12], and minimum punch diameters of 15 μm can also be achieved [65] (see Fig. 5.34b). With the use of lithography-etching processes even smaller structures can be realized.

Reference [65] also presents a possibility for the improvement of tool alignment accuracy by using an optical image acquisition system combining images of the punch tip end and the die hole. For this, a cross correlation image processing algorithm which measures the similarity between two images was used. The actual tool alignment could be done by using a movable xy-stepper adjusting the horizontal position of the die hole under the punch tip. However, the accuracy of this tool alignment method was highly influenced by the edge sharpness of the punch tip. As a result tool alignment accuracy was limited to an eccentricity error of about 1 μm .



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Fig. 5.34 Tungsten carbide punches with **a** a diameter of 25 μm , fabricated by ultra-precision micro grinding [32], **b** a diameter of 15 μm , fabricated by wire-EDM [65] (with kind permission of Springer Science + Business Media)

Table 5.4 Effect of tool material and tool coating on punch wear [6]

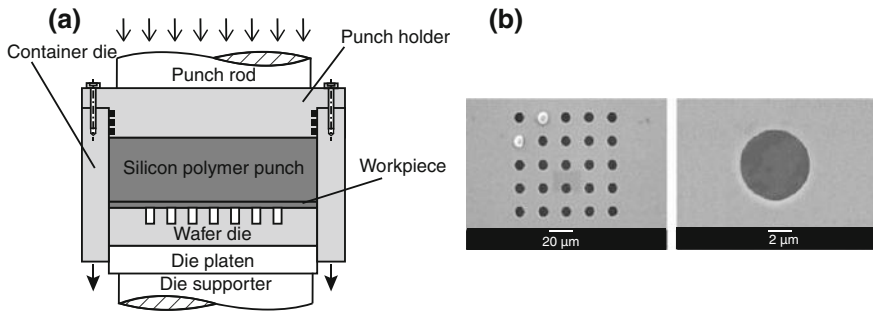
Tool material and coating	Flank wear (μm)	Face wear (μm)
HSS	35.0	37.8
TiN coating on HSS	32.8	20.0
WC	17.8	16.7
Plasma-nitrided HSS	30.6	15.0

Another important factor in micro piercing and punching is the occurrence of tool wear. In [6] a comparative investigation on the wear tendency of different tool materials and coatings for the dam-bar cutting of integrated circuit packages was carried out. There, tools made of high speed steel exhibited the poorest wear resistance while a titanium nitride coating by PVD and a surface treatment by plasma nitriding reduced the run-in wear at the cutting edge and the flank of the punch (see Table 5.4). Tungsten carbide has been proved to have the best wear resistance and is the most commonly used tool material in piercing and blanking [6].

However, even if increased efforts have been made to realize piercing and blanking in micro scale dimensions there are some restrictions left, complicating the trouble-free execution of the process. For example, even if it is possible to achieve the necessary accuracy in manufacturing, assembly and alignment of the tools, the dynamic behavior of machines and tooling and the resulting deflections are very likely to cause an offset of the punch or the die of more than 1–2 μm . Consequently resulting in damage to the tools and increased wear. If higher clearances than appropriate are used burr formation may not be avoidable, which would result in the need for an additional process step to remove them, since they are unacceptable for further use in e.g. electronic products [47]. Also it might be necessary to perform a conventional micro piercing or blanking process in a dust-free and temperature-controlled environment, especially when multiple punches are used to satisfy the precision requirements [31].

5.5.3 Alternative Micro Piercing and Blanking Processes

To circumvent some of the previously-mentioned problems of a downscaled conventional piercing or blanking processes using rigid tools, some alternative technologies were investigated. For instance, in [48] a flexible polymer tool for the simultaneous piercing of multiple holes instead of metal punches has been proposed. This method is similar to rubber pad forming which is already applied in conventional sheet metal punching. Figure 5.35 shows the concept of the flexible punching process, as well as the resulting product. The major tools in this process are the silicon polymer punch and the corresponding die. In this setup, made from a single crystal Si wafer with 500 μm thickness, the workpiece foil is positioned between the wafer die and polymer punch and all these parts are placed inside the container die and are crowned by the flat punch holder which is necessary to



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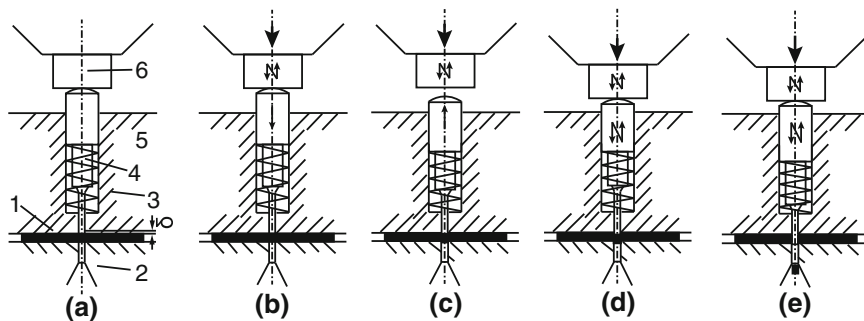
Fig. 5.35 a Schematic of the flexible punching process [48]. b 8 μm-diameter holes punched on titanium foil with a thickness of 1.5 μm [48] (reprinted with permission from Elsevier)

transmit the punch motion to the polymer punch. To prevent the silicon polymer from squeezing out through the gap between the container, die and blank holder seal rings are used.

Experiments on the feasibility of this method as an operational piercing process were conducted by Rhim et al. [48] using rolled pure copper of 3 μm in thickness and CP titanium of 1.5 μm in thickness as foil material. The diameters of the manufactured holes ranged from 2 to 10 μm. For the experiments, dies with circular as well as rectangular holes were used under non-lubricated conditions. The results of this investigation show that a piercing process using a flexible punch is suited for production of very small holes, unobtainable by conventional piercing, down to a diameter of 2 μm. The workpiece foil was easily separated from the punch and die after piercing, without excessive force for separation. However, the process of die fabrication needs to be improved, since the etching process used provides some unwanted restrictions.

Another alternative to conventional piercing is presented by Takemasu et al. [57], where a simple and effective piercing system using ultrasonic vibration was newly developed for the production of micro holes with fine sheared surfaces. The new process, named shuttle piercing, consists of two independent units: an ultrasonic vibration unit and a piercing unit. A schematic representation of the process is given in Fig. 5.36. In the piercing unit, a coil spring is placed between the punch and the guide bush to support the punch in a floating state. After inserting the target sheet or foil and its fixation between the die and the guide bush, the spring-supported punch is moved downwards with a constant velocity while the ultrasonic vibration horn (UV-horn) simultaneously generates a longitudinal vibration. Hereby, a back-and-forth shuttling motion of the punch is continuously performed and the piercing process is carried out incrementally until a hole is pierced.

Investigations by Takemasu et al. [57] using commercial thin steel sheet with thicknesses ranging from 0.08 to 0.15 mm and a tungsten carbide punch with a tip diameter of 0.13 mm, demonstrated the technological feasibility and the advantages of this process. The constant feed rate was 0.15 mm/s and an ultrasonic



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Fig. 5.36 Schematic of shuttle piercing [57]. 1 Material sheet, 2 Die, 3 Guide-bush, 4 Coil spring, 5 Punch, 6 U V-horn

vibration frequency of 40 kHz was applied. The results achieved were compared with those of a conventional piercing process. While the conventional process showed a little rollover near the entrance of the pierced holes and the percentage ratio of the fractured zone to the sheet thickness was 20–25 %, shuttle piercing results revealed that the heights of the fractured zone were significantly diminished or almost eliminated. The percentage ratio of the fractured zone was 5 % or less in every case, and it was confirmed that the punch shuttling motion suppresses the occurrence and the growth of destructive cracks. Not only vertical holes but also inclined holes were pierced. Here, conventional piercing resulted in a large roll-over and the fractured zone covered about 15 % of the sheet thickness. In shuttle piercing rollover could not be detected, the sheared surface was smooth and the fractured zone almost disappeared. In summary, it can be emphasized that this process shows great potential for high precision piercing, although the process velocity will remain below that of conventional piercing.

5.5.4 Influence of Miniaturization on the Piercing and Blanking Process

One of the first studies regarding the systematic investigation of the effects of miniaturization on a piercing and blanking process was done by Geiger et al. [18]. Starting with a sheet thickness of 1 mm and a punch diameter of 20 mm, the process was scaled down with factors of 0.5, 0.2 and 0.1. The maximum punching forces achieved from the experiment were compared to the calculated values of the same scaling factor. While for a scaling factor of 0.5 the punching force was predicted very precisely, further miniaturization led to a strong increase in the percentage deviation. The reason for this was anticipated as being the concentration of the shear deformation in just a few grains. Due to their limited number of

sliding planes and their constrained position they may have caused a larger shearing resistance than in a polycrystalline material with its large number of grains and grain boundaries.

Another study, investigating the influence of miniaturization on the piercing and blanking process, was conducted by Joo et al. [32]. The piercing of holes with a thickness to diameter ratio of 1 was carried out for 100, 50 and 25 μm foils of brass and stainless steel. It was found that the quality of the punched holes, as well as the shape of the sheared edge, changed with miniaturization. A decrease in the hole diameter resulted in a smaller part of the fracture depth and an increase in burnish depth and burr height (see Table 5.5). The mechanism of ductile fracture by crack initiation and propagation, as typically known from conventional piercing, could be found when the piercing of the large holes was investigated. As a consequence, a clearly visible fractured area in the sheared edge was the result. With increasing miniaturization, this mechanism did not seem to work. Rather, shear deformation was dominant but no crack initiation and propagation could be found until shortly before the end of the piercing process. Hereby, the breakout diameter met the punch diameter better than did the piercing of larger holes, although the burr height increased. As an explanation for this difference in behavior, the number of grains over the foil thickness was assumed to be the reason.

Reference [36] also investigated the influence of miniaturization on the sheared edge, and revealed as one characteristic effects of miniaturization an increasing burr height. Furthermore, miniaturization led to an increasing irregular development of the sheared edge, where the orientation of single grains towards the shearing direction seemed to become rather decisive. The same specimen could show extremely different sheared edges, making it difficult to quantify the parts of burnish depth and fracture depth when the downscaled process was considered.

Reference [11] conducted blanking tests, developed in order to simulate a part of a leadframe manufacturing process, carried out with a long and very narrow punch and FeNi42 material in different thicknesses, which revealed an increased influence of the material's anisotropy. Since the punch is much longer than it is wide, nearly 95 % of the shearing line had the same orientation which results in a punch force dependency on the rolling direction of the sheet material. Furthermore, the burnish depth of the sheared edge was also found to be a function of the punch orientation.

Table 5.5 Influence of miniaturization on the sheared edge for brass and AISI 316 [32]

Foil material	Brass	Brass	Brass	AISI 316	AISI 316	AISI 316
Foil thickness (μm)	100	50	25	100	50	25
Tool size (μm)	100	50	25	100	50	25
Rollover depth (μm)	4.7	1.3	–	1.1	3.3	–
Burnish depth (μm)	91.4	42.9	25	77.7	46.6	25
Fracture depth (μm)	8.6	5.7	–	22.2	–	–
Burr height/thickness (%)	<1.4	<2	<6	<1.5	<4	<4

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Chapter 6

Size Effect-Enabled Methods

Heinz-Rolf Stock

Latin

c_p	Heat capacity (J/(kg K))
d	Diameter (mm)
ES	Surface energy (J)
F	Force (N)
k_f	Yield stress (MPa)
l	Length (mm)
P	Power (W)
Q	Heat (J)
q	Latent heat (J/kg)
r	Radius (mm)
t	Time (s)
T	Temperature [$^{\circ}\text{C}$ (K)]
T_D	Deposition temperature [$^{\circ}\text{C}$ (K)]
T_M	Melting temperature [$^{\circ}\text{C}$ (K)]
u	Upset ratio (-)
v	Velocity (mm/s)
V	Volume (m^3)

Greek

α	Absorption coefficient (-)
γ	Surface tension (N/m)
η	Efficiency (-)
Φ	Amplitude (-)
	Divergence angle ($^{\circ}$)

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λ	Wavelength (m)
ρ	Density (kg/m^3)
θ	Focus diameter (mm)
σ_M	Tensile strength (MPa)
σ_N	Nominal stress (MPa)
φ	Natural strain (-)

6.1 Additive Sheet Generation

Heinz-Rolf Stock and Helge Decho

6.1.1 Thin Sheet Generation: Common Methods and Alternatives

Micro metal forming requires rather thin parts as the wrought material. Typical thicknesses for sheets are in the range of 10–50 μm . Such sheets can be produced by cold rolling. Common materials, which can be rolled down to such low thicknesses, are for example aluminum (Al99.5), low carbon steel, stainless steel or copper (E58). These materials are available in large quantities at moderate prices. Aluminum foil is for example not only used for the packaging of food but also for thermal insulation, cables and electronics. These materials are work hardened during the cold rolling process and are annealed afterwards to recover their original material properties. All these materials mentioned are rather soft and pliable. This diminishes their potential as construction materials for micro parts.

Other materials, like hardened tool steels or high-strength aluminum alloys in the form of thin sheets, would be more useful for micro metal forming, but—to our knowledge—are not available in such low thicknesses. In principle, it is possible to roll each material down to several 10 μm . Demura et al. for example produced 23 μm thin foils of Ni_3Al starting from 5 mm thick single crystals [8]. But this process needs ambitious machining, e.g., a Sendzimir planetary mill [24] and accurate annealing conditions to prevent cracking and tearing. It is therefore worth considering alternative processes for the production of thin sheets. Potential candidates for this purpose are electroplating, thermal spraying and vapor deposition. Balancing the options of these three candidates, electroplating fails because only few metals like chromium or nickel can be deposited, but neither tool steels nor high-strength aluminum alloys. Thermal spraying can be used to produce high-strength materials, but voids within the deposited material cannot be avoided and so drastically decreases the mechanical properties.

The most promising alternative is therefore vapor deposition, especially physical vapor deposition (PVD). This method implies the evaporation of the source material, transport of the gaseous atoms and deposition, atom by atom, on a

substrate. This substrate has to afterwards be removed from the deposited foil. There are several options to evaporate the source material; among them sputtering is the method which will be reported here.

6.1.2 Sputtering: Principles

In physical vapor deposition processes coatings are formed from a flux of atoms that approaches the substrate from an evaporated source. It consists of three steps:

- Evaporation of a desired material, resulting in a flux of particles like atoms, ions or molecules,
- transportation of the particles towards the substrate, and
- condensation of the particles on the substrate.

With regard to step one there are five methods to evaporate materials: Thermal evaporation, electron beam evaporation, arc evaporation, laser evaporation and sputtering. Among them sputtering has the advantage that it is not expensive and alloys can be evaporated—and afterwards recondensed—without significantly changing their chemical composition.

The first step of a sputtering process is the ignition of a glow discharge. The phenomena to be observed in a direct current (DC) glow discharge are shown schematically in Fig. 6.1 [7]. In a vacuum tube filled with argon (up to 1,000 Pa) two electrodes are placed and a voltage of several hundred volts is applied between them. Within a few moments after applying the voltage some random argon ions are accelerated towards the cathode, and on its way generate several other ions by collision—resulting in a self-sustaining glow discharge. This physical condition is called a plasma. Different glowing areas and dark spaces between them can now be observed.

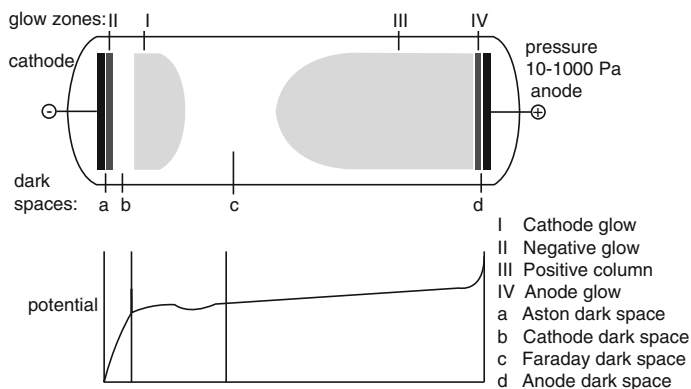


Fig. 6.1 Arrangement of a DC glow discharge, visible glow zones and dark spaces, and corresponding potential gradient

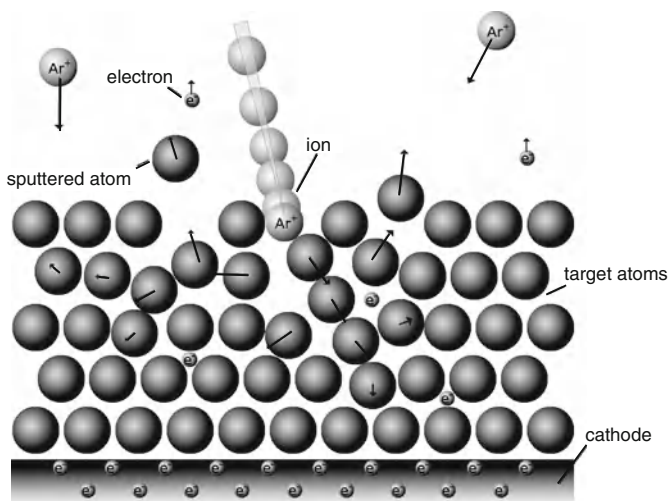


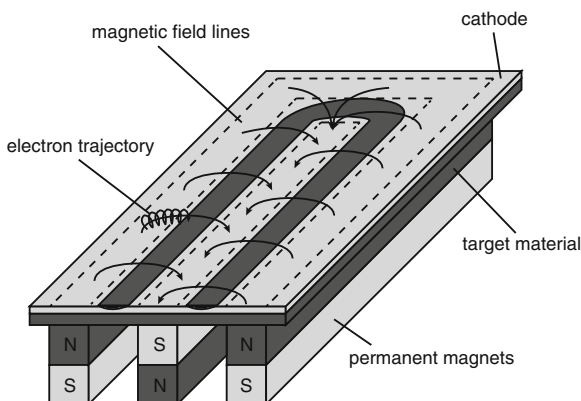
Fig. 6.2 Schematic of a sputtering incident: an incoming argon ion hits the target surface and causes target atoms to be knocked free

The argon ions are driven towards the cathode and finally hit the surface of the cathode. Depending on their energy these ions cause atoms—and even clusters of atoms—to be knocked free from the cathode surface by momentum transfer, shown schematically in Fig. 6.2. Due to this bombarding process the cathode material is usually called the *target*.

These atoms have a certain amount of kinetic energy and move away from the target in a straight line. To prevent collisions with gas atom the pressure in the vacuum chamber must be in the range of 0.1–1 Pa. With that the mean free path is in the range of several centimeters and allows the majority of the atoms to reach the substrate without any collisions with gas atoms or molecules. This straightforward movement of sputtered atoms leads to one significant drawback: Sputtering is a line-in-sight process, in other words only those areas which can “see” the sputtering source are coated. To overcome this difficulty special efforts have to be made, e.g. the use of more than one source or the rotation of the substrate.

A second disadvantage is that such low pressure reduces the plasma intensity, and with that the bombardment of the target with ions and the flux of atoms away from the target. To overcome this disadvantage the flux can be significantly enhanced using magnets—thus upgrading the sputtering to magnetron sputtering. Some strong permanent magnets placed behind the target—as depicted in Fig. 6.3—generate magnetic fields. These fields have no effect on the atoms and only a minor influence on the relatively heavy ions, but strongly influence the light electrons. These electrons—they stem from the ionization processes or directly from the negatively loaded cathode (secondary electrons)—are forced into trajectories of convoluted spiral-like patterns across the target surface. This high electron density near the target surface increases the ionization probability and

Fig. 6.3 Arrangement of target and permanent magnets in a magnetron sputtering unit. Magnetic field lines force electrons into circular trajectories and with that enhance the ionization process along that path



thus raises the amount of sputtered target material and the resulting deposition rate by more than an order of magnitude.

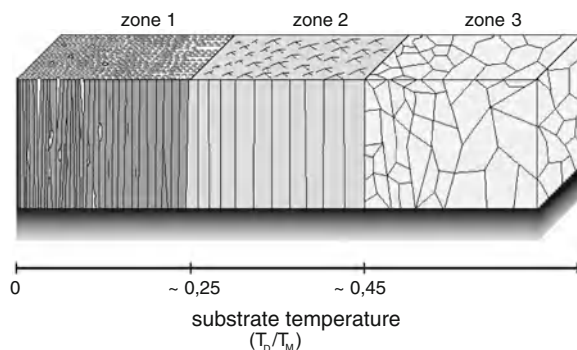
One well-established method to enhance the mechanical properties of the growing coatings, especially the adhesion, density and morphology, is the application of a so-called bias voltage. This means that a negative potential—in the range of -100 V—is applied to the substrates. This causes a slight bombardment of the growing film which results in the sputtering of already deposited particles, especially those with a reduced bonding to the surface. As a consequence the growing film becomes denser and a better adhesion to the substrates is achieved. Furthermore, due to the sputtering of the growing film the overall deposition rate is reduced.

In addition there are several possibilities for enhancing the plasma intensity, either by intensifying the magnetic field in the center of the vacuum chamber (known as *unbalanced magnetron sputtering*) or intensifying the glow discharge by means of a pulsed voltage. Latest developments in this topic are called *high pulse power magnetron sputtering* (HPPMS) or *high power impulse magnetron sputtering* (HIPIMS) with current densities up to 10^6 Acm $^{-2}$ [18].

6.1.3 Morphology of Sputtered Foils

Depending on the preparation conditions, the microstructures of thin films prepared by physical vapor deposition show a wide range. This has led to the development of models, which relate the microstructure of a film to the deposition parameters. Movchan and Demchishin were the first to investigate the influence of the substrate temperature [15]. They concluded that the structure of thermally evaporated films could be described as a function of the quotient of the deposition temperature and melting temperature of the film's material $[T_D/T_M]$, both expressed in degrees Kelvin. They distinguished three zones as shown in Fig. 6.4. Zone 1 (at $T_D/T_M < 0.3$) show tapered crystals with domed tops which increase in

Fig. 6.4 Structure zone model according to Movchan and Demchishin [15]



width with temperature. Zone 2 ($0.3 \leq T_D/T_M \leq 0.5$) consists of columnar grains with smooth matt surfaces. Zone 3 ($T_D/T_M > 0.5$) show equiaxed grains and a bright surface.

The development of these zones can be explained by diffusion. Zone 1 is determined by low adatom mobility—within the meaning of *no diffusion*. As the temperature is increased, surface diffusion becomes more important and leads to parallel sided columnar grains (zone 2). With a further increase in temperature bulk diffusion also becomes important, resulting in equiaxed grains in zone 3.

This model was extended by Thornton [19] from thermal evaporation to sputtering by the addition of a third axis to account for sputtering gas pressure in the range of 0.1–4 Pa, as shown in Fig. 6.5. The main extension of Thornton's model is a fourth zone—named the *transition zone* (zone T)—which emerges in-between zones 1 and 2 as the pressure of the sputtering gas is decreased. This zone T consists of poorly-defined fibrous grains. For many applications—like wear-resistant nitride coatings, e.g., titanium nitride (TiN), titanium-aluminum nitride (TiAlN) and chromium nitride (CrN) on high speed steel (HSS)—the morphology of zone T reveals the best wear resistance, whereas the mechanical properties of zone 1 are poor. Zone 2 coatings are also promising, but the deposition temperature required is too high. As the melting temperature of these

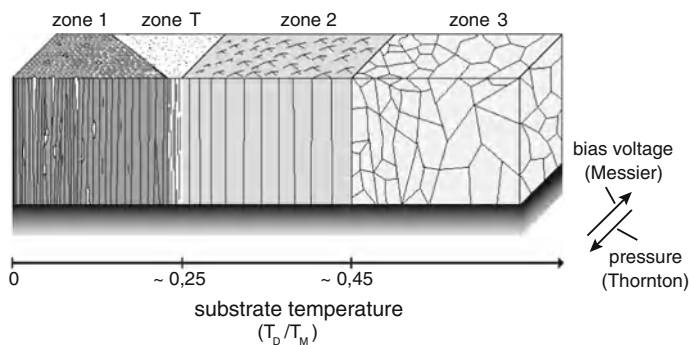


Fig. 6.5 Structure zone model according to Thornton [19] and Messier et al. [13]

nitrides are high (for TiN $T_M \approx 3,000$ °C) and the substrate temperature during deposition must not exceed the annealing temperature of the HSS (in most cases 550 °C) T_D/T_M must not exceed 0.25.

As an alternative to decreasing the deposition pressure, zone 1 can be prevented by the bombardment of the growing coating with particles of sufficient energy. This causes the filling of the voided boundaries by coating atoms, leading to a structure type of zone T [6]. This bombardment can be achieved by a bias voltage applied to the substrates (as mentioned above). This causes knock-on sputtering of already deposited atoms and induces the enhanced mobility of these atoms. These advantages were integrated into Thornton's model by Messier et al. [13] by just modifying the third axis, as already shown in Fig. 6.5.

Up to now, further ameliorations of the structure zone models are occurring. They focus upon the incorporation of intensified sputtering methods like *closed-field unbalanced magnetron sputtering* [11] or on other parameters like the segregation of impurities [2].

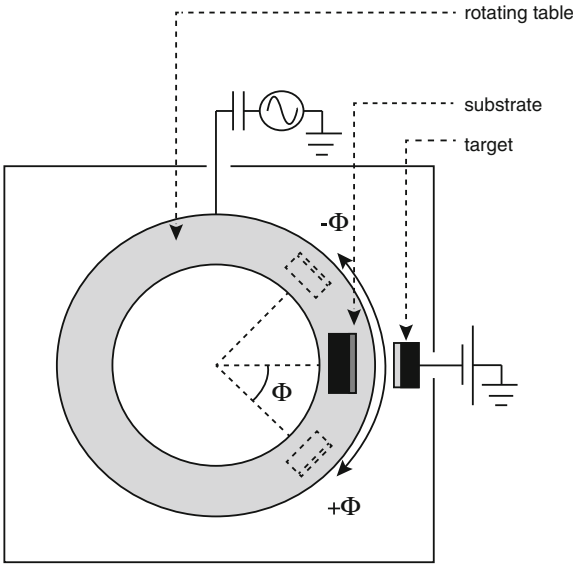
6.1.4 Sputtering of High Strength Aluminum Foils

As an example we will take a closer look at a magnetron sputtered aluminum alloy doped with scandium. When added to aluminum, scandium acts as a grain refiner and recrystallization inhibitor. The high hardness and high strength of aluminum-scandium alloys are caused by the precipitation of the intermetallic phase Al_3Sc , which is coherent with the aluminum matrix. The coherency mismatch of about 1.2 % results in significant lattice stress, which hinders dislocation motion.

The deposition took place in a magnetron sputtering unit with DC power supply, schematically shown in Fig. 6.6. An aluminum-scandium pre-alloy with a scandium content of 2.0 mass % was used as the target material, and 100 μm thick sheets of unalloyed steel were used as substrates. The latter were mounted on a temperature-controlled copper block in front of the target. The distance between substrate and target was 50 mm. To start the sputtering process the chamber was filled with argon up to a pressure of 0.6 Pa and a voltage of about -500 V was applied to the targets. Thus, a self-sustaining glow discharge is generated. The applied voltage was controlled to keep the target power constant at either 1 or 2 kW during deposition. The substrate temperature was kept constant at either 310 or 433 K by heating or cooling the copper block throughout the deposition time. Temperature was controlled with a thermocouple mounted between the copper block and the back side of the substrate.

As mentioned previously sputtering is a line-in-sight process, therefore special efforts have to be met to produce sheets with a homogeneous thickness distribution of the coatings. This could be the use of more than one sputter source, a rotation of the substrates or a combination of both. Another or simple method is a reciprocal motion of the substrate, which was applied here. As can be seen in Fig. 6.6, the reciprocal motion can be described by its amplitude Φ which is the angle between

Fig. 6.6 Schematic of the magnetron sputtering unit, which allows the reciprocal motion of the substrate



the center of the target and the center of the substrate. Hence during the deposition process, the substrate holder consistently shifts between $-\Phi$ and $+\Phi$. Figure 6.7 shows the thickness profiles of four sheets deposited with different motion amplitudes between 0° (no motion) and 23° . In the case of a motionless substrate (0°) or a movement of small amplitude (8°), thick coatings with very uneven thickness distributions are observed. With an amplitude of 15° the maximum thickness and the thickness decline towards the edges are reduced. Finally, sheets produced with an amplitude of 23° show a relatively thin but homogeneous thickness of $8\text{ }\mu\text{m}$. Therefore, this amplitude was used in the deposition experiments described in the following.

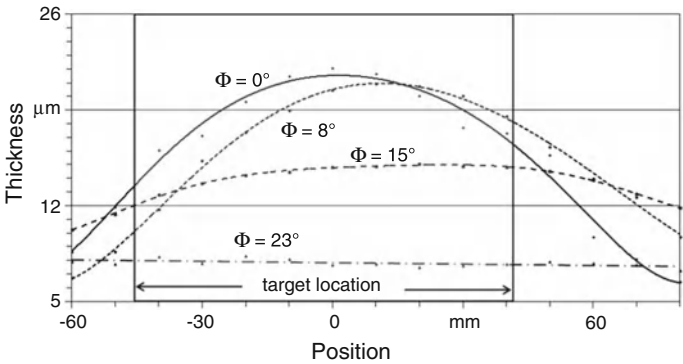


Fig. 6.7 Thickness profiles produced with different motion amplitudes

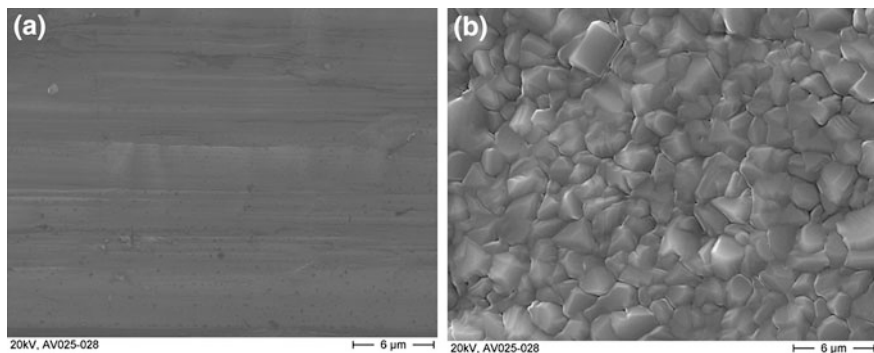


Fig. 6.8 **a** Contact surface and **b** free surface of a sputtered aluminum-scandium thin sheet

After deposition, the aluminum-scandium coatings and the steel substrates are well adhered to each other. To separate the coating from the steel substrate a nitric acid solution can be used. In this way, the steel is dissolved whereas the thin free-standing sheet of aluminum-scandium alloy remain unchanged. The outer surfaces of the separated coatings look different; we distinguish between the *contact surface* (during deposition in contact with the steel substrate) and the *free surface*, both are shown in Fig. 6.8. It is obvious that the contact surface is relatively smooth compared to the free surface. On the latter the top ends of the columnar crystals are visible. The growth of these columns can be explained by the structure zone model mentioned in Sect. 6.1.3. For the Al–Sc alloy a melting point of 655 °C (938 K) can be assumed, slightly below that of pure aluminum (660 °C). Therefore, the Al–Sc foils were deposited at $T_D/T_M = 0.33$ or 0.46 , which in both cases led to zone 2 with its typical columnar grains.

6.1.5 Mechanical Properties of High Strength Aluminum Foils

To assess the influence of artificial ageing on the mechanical properties of the aluminum-scandium sheets, three temperatures (200, 300 and 400 °C) were chosen. The holding time was 1 h and additionally 0.5 h at 400 °C. The resulting hardness values are presented in Fig. 6.9. Compared to the hardness of the untreated sheet, the hardness is increased by annealing for 1 h at 200 °C and even more after 1 h at 300 °C. The annealing at 400 °C caused a slight decrease in hardness, indicating that 1 h at 400 °C is an over ageing condition. By reducing the treatment time to 0.5 h at 400 °C this effect does not occur, and the resulting hardness remains relatively high. A second post-treatment consisted in cold isostatic pressing at 400 MPa. The aim of this treatment was to reduce the pores in the bulk of the sheets.

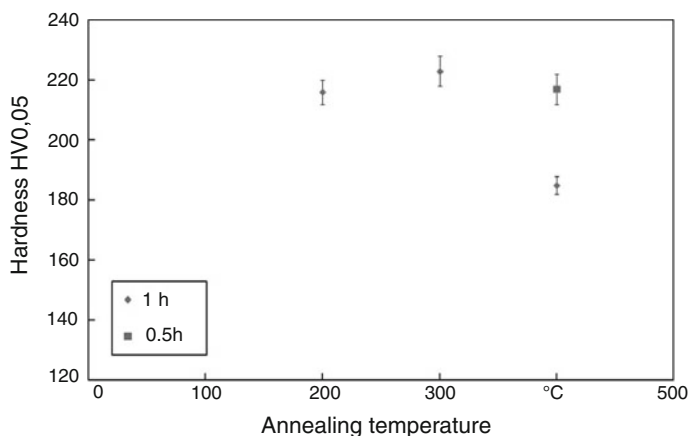


Fig. 6.9 Hardness of Al-Sc sheets after artificial ageing at different temperatures

Heat treatment to increase the hardness and strength of aluminum alloys normally consists of three steps: Firstly a solution heat treatment to dissolve the soluble phases, secondly quenching to achieve the supersaturation of alloying elements and vacancies, and thirdly age hardening to precipitate soluted supersaturated atoms. During the magnetron sputtering process supersaturation is settled by vapor depositing the alloy atom by atom, and simultaneously avoiding too much diffusion by limiting the substrate temperature. The third step—age hardening—is either achieved at room temperature or with a precipitation heat treatment, also known as artificial ageing. As reviewed by Royset and Ryum, it has been shown that Al_3Sc particles precipitating in the temperature range between 250 and 350 °C cause a significant hardness increase in the alloy [17]. This can be approved by the hardness measurements after the artificial ageing between 200 and 400 °C as depicted in Fig. 6.9. As known from other hard enable aluminum alloys, the optimum artificial ageing to achieve maximum strength depends on the temperature and annealing time [1]. The chosen ageing parameters—300 °C and 1 h—show an increase in hardness from 187 to 223 HV, both values are relatively high.

The most influential deposition parameters are target power and substrate temperature. Therefore, four deposition experiments were conducted, using the parameters listed in Table 6.1. After separation of the coatings from the steel

Table 6.1 Deposition parameters and resulting coating thicknesses

Number of deposition treatment	Substrate temperature (°C)	Target power (kW)	Deposition time (h)	Coating thickness (μm)
1	160	1	4	32
2	160	2	2	23
3	37	1	4	32
4	37	2	2	21

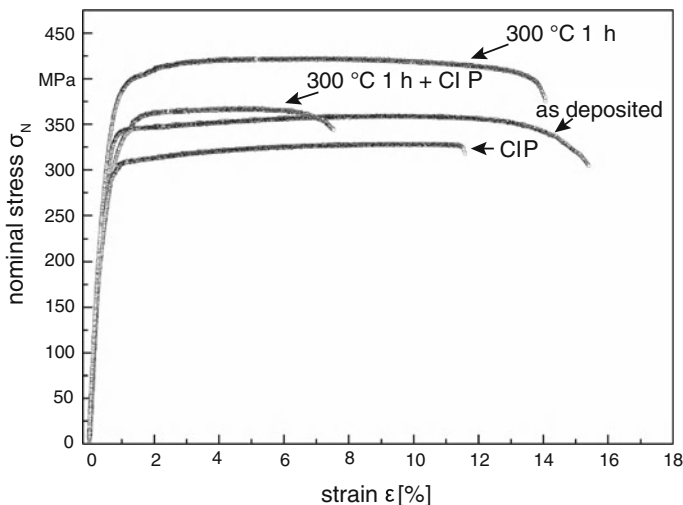


Fig. 6.10 Tensile tests on four Al–Sc foils from deposition treatment No. 1 (1 kW, 160 °C, 4 h) with different post-treatments

substrates the samples were either annealed at 300 °C or cold isostatic pressed or both. Together with the as-deposited probes, altogether 16 different samples of tensile test specimens were manufactured.

In Fig. 6.10 the tensile test curves of coating No. 1—treated at 1 kW target power at 160 °C for 4 h—and different post-treatments are shown. The Al–Sc foil No. 1A (1 kW, 433 K, 4 h, without post-treatment)—referred to “as deposited”—has a tensile strength of 351 MPa. A heat treatment at 300 °C for 1 h enhances the tensile strength to 424 MPa. In contrast with that cold isostatic pressing (CIP) reduces the tensile strength to 327 MPa, and again subsequent heat treatment increases this value to 370 MPa. Figure 6.11 graphically illustrates the results.

The mean value of all tensile strengths in Fig. 6.11 results in $\sigma_M = 349 \pm 11$ MPa, which is relatively high. Especially if it is considered that the applied tensile force acts perpendicularly to the aluminum columns. It is worthwhile taking a closer look at the influence of the deposition parameters and the post-treatments on the tensile strength σ_M . It is obvious that the deposition temperature influences the tensile strength. The mean tensile strength of all eight tests with coatings deposited at 160 °C slightly increase the σ_M to 360 ± 10 MPa—compared to the mean value $\sigma_M = 349 \pm 11$ MPa. In contrast a deposition temperature of 37 °C slightly reduces the tensile strength ($\sigma_M = 339$ MPa). Apparently the target power has no significant influence, σ_M values of 351 ± 11 and 347 ± 12 MPa are within the margin of error compared with the mean value.

Figure 6.11 also reveals the influence of the two post-treatments. The as-sputtered aluminum-scandium foils show a relatively low tensile strength ($\sigma_M = 314 \pm 19$ MPa). However, the annealing treatment for 1 h at 300 °C significantly raises the tensile strength to $\sigma_M = 405 \pm 13$ MPa. In contrast, cold isostatic

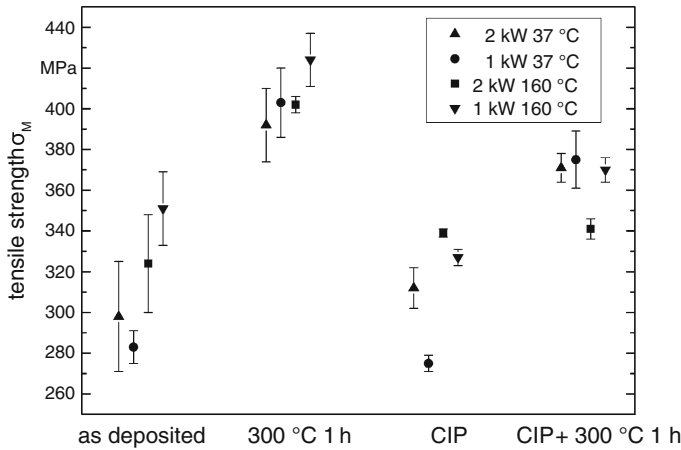


Fig. 6.11 Tensile strength values due to different deposition and post-treatments

pressing does not increase the σ_M but the standard deviation of the σ_M values achieved is significantly reduced (to ± 5 MPa). The combination of both post-treatments results in a mixture of both effects: The tensile strength (364 ± 8 MPa) is increased but not as much as with sole annealing, and the standard deviation is reduced (to ± 8 MPa) but to a lesser extent than with sole cold isostatic pressing.

6.2 Thermal Free form Heading

Heiko Brüning

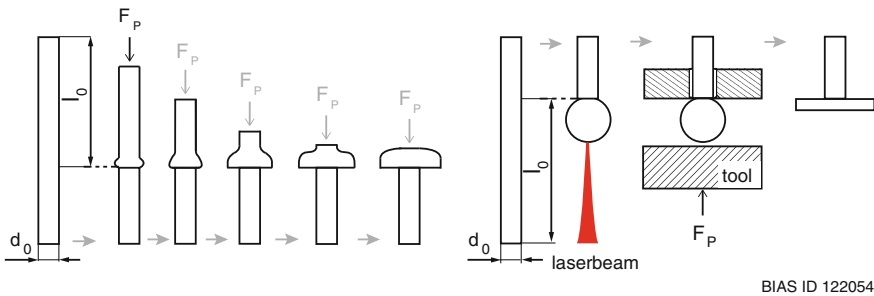
6.2.1 Basic Aspects of Thermal Free form Heading

Bulk metal forming is a process which allows the production of goods with a high accuracy in very low cycle times [9]. Due to size effects, process limits known in the macro range cannot be easily transferred into the micro range. Especially, but not exclusively, in the limit in the single step upsetting processes the upset ratio u is reduced from $u = 2.3$ [12] to a maximum of $u \leq 1.5$ for workpieces with dimensions less than 1 mm [14] (see Sect. 4.2). Greater upset ratios can only be achieved by enhancing the number of upsetting steps combined with recrystallization annealing, leading to an increase in process complexity, process time and overall costs if the conventional upsetting process is taken into account.

In order to still be able to benefit from the advantages of bulk metal forming without increasing the number of forming steps necessary, a free form upsetting process called “thermal free form heading” has been designed and investigated. The process consists of two main steps: the first generates a material accumulation by laser melting which is afterwards calibrated in a second step, e.g. by upsetting.

6.2.2 Process Principle and Variants

Cylindrical wrought materials, such as rods, are used as the basis for the thermal free form heading process. The longitudinal axis of the rod is oriented parallel to the gravitational force. A heat source causes the end of the rod to melt. Within certain limits, the molten material sticks to the rod and forms a sphere which is called “material accumulation” or “preform” if a forming process is attached. It is advisable to use a laser beam as the heat source so that the amount of energy can be adjusted accurately in terms of amount and time such that it leads to a robust process. Generating material accumulations can either take place at the bottom end of a rod or at the top end, relative to the direction of gravitational force. If the process is carried out on the top surface it is called “upright position”, the opposite arrangement is called the “hanging position”. Figure 6.12 illustrates both, the process of conventional upsetting as well as generating preforms by thermal free form heading in a hanging position with an attached open die upsetting step. The upset ratio u is defined as the ratio of accumulated or upset length l_0 and the initial rod diameter d_0 . F_P represents the required punch force for each forming operation. It is to be noted that the thermal free form heading process is supposed to take place under inert gas flow conditions if wrought material susceptible to oxidation at high temperatures is used.

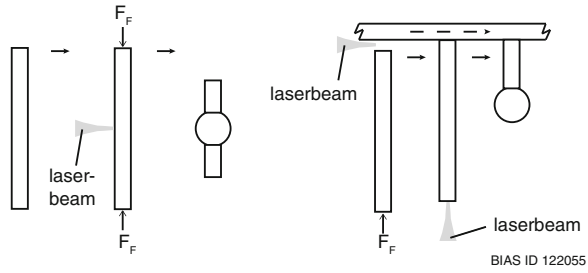


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Fig. 6.12 *Left* Conventional upsetting process. *Right* Upsetting with preform generated by thermal free form heading in hanging position

The smaller the semi-finished products are the more difficult positioning becomes. This challenge is omnipresent in the micro range and is amplified by the fact that forces related to surface tension reach or even exceed the magnitude of gravitational forces, so that the handling of goods is also a difficulty. Concerning the topic of material accumulations, handling is simplified by producing continuously conveyable structures. In this case, the process shown in Fig. 6.12 is modified to enable either the generation of material accumulation within a continuous rod by applying a feeding force F_F or the fixing of the shaft of the material accumulation to a carrier rod. These variants are schematically shown in Fig. 6.13.

Fig. 6.13 Continuously conveyable structures. *Left* Material accumulation within a continuous rod. *Right* Material accumulation and carrier rod



6.2.3 Process Analysis

6.2.3.1 Process Limits

Thermal free form heading is a process which is enabled by effect of surface tension becoming more and more dominant compared to gravitational forces in the micro range. The melt at the end of the rod does not release as long as the surface tension at the border between liquid and solid material exceeds the volume forces due to gravitation. Beyond this limit the material accumulation detaches. The ratio of rod diameter d_0 and effective upsetting length l_0 , called the upset ratio u , can be calculated for generating material accumulations in hanging positions assuming that the droplet forms a perfect sphere and the intersectional area between rod and melt is even and circular. As shown in Fig. 6.14, two main forces act on the droplet: the gravitational force F_G and the surface tension F_γ at the intersectional area between the melt and rod [20].

$$F_\gamma = \pi d_0 \gamma_0 \quad (6.1)$$

$$F_G = \frac{1}{6} \pi d_s^3 \rho g \quad (6.2)$$

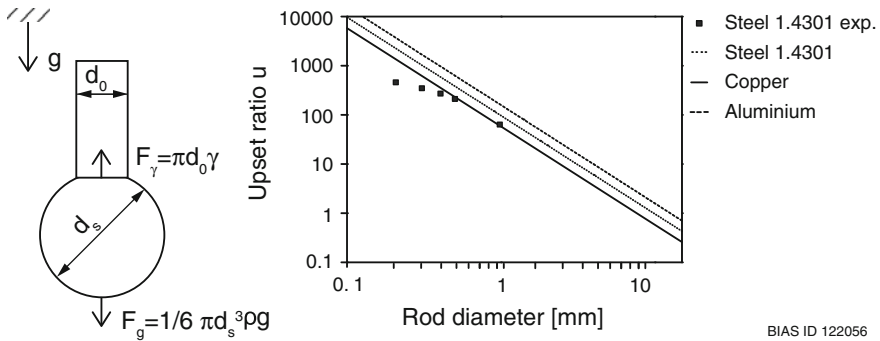


Fig. 6.14 *Left* Material accumulation in hanging position with acting forces. *Right* Maximum achievable upset ratios dependence on rod diameter and material

The maximum achievable upset ratio u is achieved if the gravitational force of material accumulation equals surface tension. This value is dependent on the density ρ , the surface tension γ_0 and the gravitation g and increases with decreasing rod diameter d_0 by the power of 2.

$$u = \frac{l_0}{d_0} = \frac{4V_S}{\pi d_0^3} = \frac{\pi d_0 \gamma_0}{\frac{1}{4} \pi d_0^3 \rho g} = \frac{4\gamma_0}{d_0^2 \rho g} \quad (6.3)$$

Formula 6.3 states that Al99.5 rod diameters have to be below 13 mm to allow an effective upsetting ratio, for steel (1.4301) below 10 mm, and for E-Cu58 below 8 mm. For rod diameters below 1.0 mm, upsetting ratios of several 100 can be achieved. Experiments with 1.4301 as the sample material show a good accordance to calculated values for maximum upset ratios for rod diameters $0.4 < d_0 < 1.0$ mm. The larger the value of the upset ratio the longer the upset length, which makes it more likely for the inert gas fluid flow to turn from laminar to turbulent. Forces caused by irregular detaching eddies initiate the rod to vibrate which leads the material accumulation to release so that theoretical upset ratios cannot be achieved for rod diameters $d_0 < 0.4$ mm unless vibration of rod is avoided. Nevertheless, upset ratios $u = 500$ for $d_0 = 0.2$ mm are reached (see Fig. 6.14). This value is far beyond the upset ratio $u \leq 1.5$ which can be achieved by the conventional upsetting process [14].

6.2.3.2 Adiabatic Energy Model

The amount of energy E_S needed to generate a spherical material accumulation with volume V_S can be estimated by an adiabatic model. E_S is the thermal energy brought into the rod to cause an increase in temperature from T_0 to the melting temperature T_M .

$$E_S = V_S \rho (c_p (T_M - T_0) + H_M) + Q_{\text{loss}} \quad (6.4)$$

Herein Q_{loss} includes among others heat conduction within the rod, convective flow of heat to inert gas, heat radiation or sputter formation.

The amount of absorbed energy E_S at normal incidence on an opaque surface is expressed by the effective absorption α_{eff} and the laser beam energy E

$$E_S = \alpha_{\text{eff}} E. \quad (6.5)$$

Setting $Q_{\text{loss}} = 0$, experiments show that this basic model still characterizes well the generation of material accumulations at the end of a rod, as long as the beam quality is rather high so that the defocusing effect does not come into account, and no abnormal absorption takes place nor keyhole exists (Fig. 6.15) [21]. The smaller the rod diameter, the better it fits the adiabatic model. This behavior is strengthened by the fact that energy dissipation due to heat transfer within the rod decreases by a square factor of the rod diameter [3]. Basic conditions to diminish the keyhole effect apply, if the power density at the sample

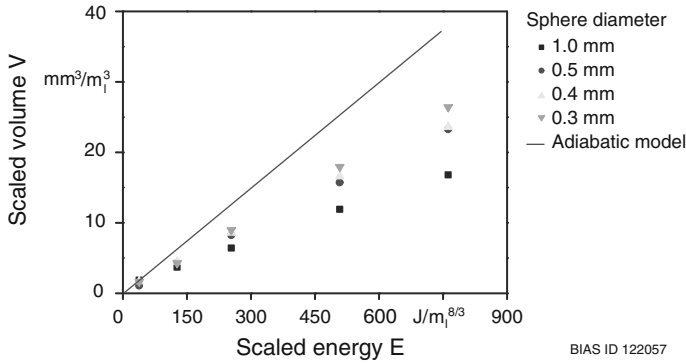


Fig. 6.15 Scaled volume dependence on scaled energy. Adiabatic model is calculated with constant absorption coefficient $\alpha_{\text{eff}} = 0.38$ [4] and $Q_{\text{loss}} = 0$ during irradiation time

surface is less than several MW/cm^2 . The defocusing effect takes place during the accumulation process as long as the focus level and the rod are kept fixed, because the beam diameter at the sample surface increases and therefore results in a decrease in absorbed energy. In contrast, only the increase in diameter of the sphere increases the amount of absorbed energy as the angle of incidence differs less intensely from 0° for larger material accumulations. The result of the defocusing effect can be seen in Fig. 6.15 as the overestimation in predicted volume by the adiabatic model, and the measured volume increases with increasing accumulated volume.

6.2.3.3 Influence of Laser Beam Source

Poprawe [16] states that generally laser material processing is not only influenced by the material parameters, which is true in case of thermal free form heading the rod (surface tension, density), but also by the interaction of the processed material and the laser beam, which is, assuming a constant wave length, influenced by:

- optical constants of the material
- chemical condition of the surface (oxidation, ...)
- geometrical condition of the surface (impurities, striations, ...)
- temperature, which also affects the above-stated parameters
- angle of incidence.

The results of the adiabatic model shown in Fig. 6.15 not only neglect heat dissipation but also the defocusing effect, so that a constant absorption coefficient during complete irradiation time is applied which does not fit the experiment. The process efficiency of the thermal free form heading process can be defined as the quotient between the energy being absorbed by the surface of the material accumulation and the total possible absorbable laser beam energy.

$$\eta = \frac{E_{\text{absorbed}}}{E_{\text{beam}} \alpha_{\text{eff}}} \tag{6.6}$$

The beam characteristic has a strong influence on process efficiency during thermal free form heading if both the focus of the laser beam and rod are kept fixed, because the developing material accumulation moves out of focus. The greater the divergence angle θ , the stronger the decrease in process efficiency.

Bearing in mind that the distance between the spherical surface of the material accumulation and the focus level is rising during the accumulation process, a small angle of divergence θ is aspired. Power densities in the focus layer exceeding several MW/cm² [21] result in the formation of a keyhole in the melted surface of the rod so that the absorption coefficient increases due to multi-scattering, so a small laser beam diameter is another objective [22]. The geometrical parameters of a laser beam in accordance with a rod and resulting material accumulation are shown in Fig. 6.16, as well as beam parameters for a fiber laser and Nd-YAG laser. The efficiency η of the process, in dependence on distance between sample surface and focus layer h , as well as the divergence angle θ is illustrated in Fig. 6.16 right for rod diameters of $d_0 = 0.3$ mm and $d_0 = 0.5$ mm.

It can be seen that a small divergence angle θ (e.g., fiber laser) enables process efficiencies η close to 1 for rod diameters $d_0 = 0.3$ mm and $d_0 = 0.5$ mm and a distance between sample and focus $h < 30$ mm, while using the Nd-YAG laser with divergence angle $\theta = 23.0^\circ$ forces the efficiency to decrease rapidly down to ≤ 10 %.

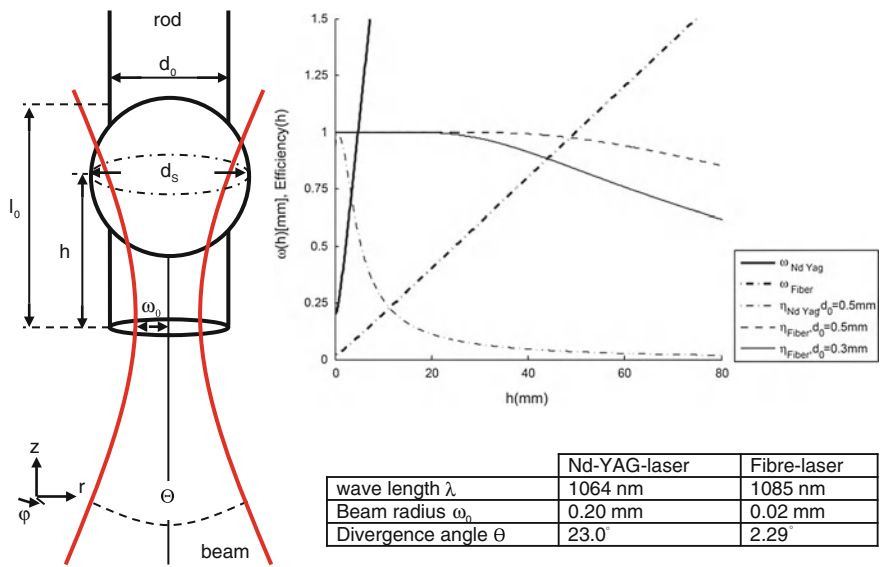
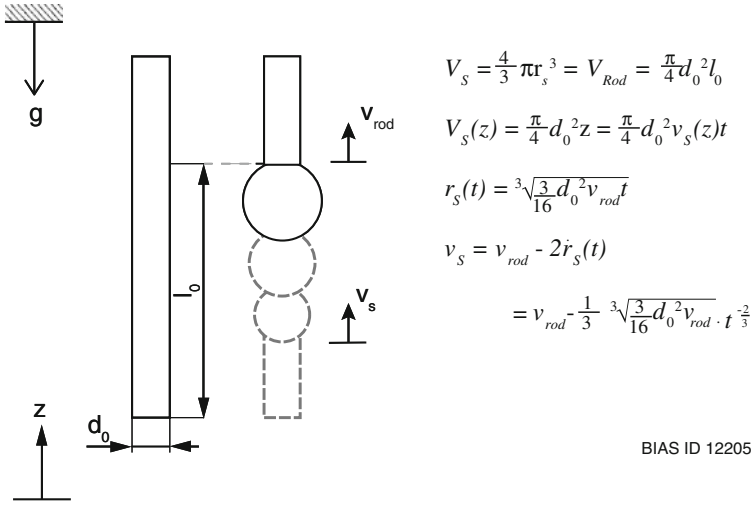


Fig. 6.16 Beam characteristic with material accumulation in hanging position and resulting process efficiency



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Fig. 6.17 Velocity of material accumulation during accumulation process

The defocusing effect can be avoided by keeping the distance between sample surface and focus constant, thus leading to an almost invariant input of energy at the surface of the material accumulation. This is achieved by either shifting the focus of the beam or by realizing feeding of the rod. In both cases, the velocity of shifting has to equal the up-moving velocity of the material accumulation for optimal conditions. Besides an increase in process efficiency, a further advantage of feeding is that laser beam quality can be lower so that initial costs are reduced. As long as a constant absorption coefficient is assumed and a constant laser beam power is applied, the vertical velocity v_s of the material accumulation surface is not constant. This is due to the fact that the volume of the sphere is growing linearly with irradiation time, but concurrently expands in all three dimensions. Independently of irradiation time, there is always the same amount of rod material liquefied per time unit; in other words the velocity v_{rod} is only dependent on the process variables of rod diameter d_0 and beam power P_L . Derivation of formula 6.4 with respect to t gives the following correlation between P_L and v_{rod} , where Q_{loss} is neglected:

$$P_L = \rho \dot{V}_s (c_p (T_M - T_0) + H_M) = \rho \frac{\pi d_0^2 v_{rod}}{4} (c_p (T_M - T_0) + H_M) \quad (6.7)$$

Based on formula 6.7, Fig. 6.17 illustrates the dependency between the velocity of the surface of the sphere v_s and v_{rod} :

Figure 6.18 shows as a result of the experiments that the energetic process efficiency η can almost be doubled by adjustment to an appropriate feeding speed. In this case a constant feeding speed was applied for simplification. The dashed line represents the upset length l_0 according to the adiabatic model, neglecting heat dissipation ($Q_{loss} = 0$) and the defocusing effect (formula 6.2.4.). For upset ratios

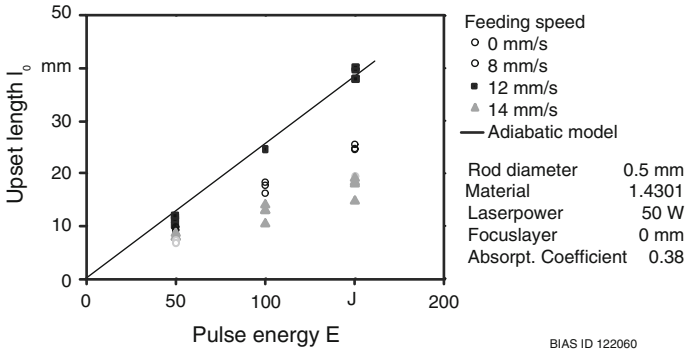


Fig. 6.18 Upset length with and without feeding compared to adiabatic model with $\alpha_{\text{eff}} = 0.38$ and $Q_{\text{loss}} = 0$

u less than 20, the dependency between the average vertical velocity of the surface of the material accumulation v_s and the power of the laser beam P_L can be stated as linear. Regarding this, high laser beam power leads to an increase in process efficiency η , which is not surprising because a lower process cycle time allows less time for heat dissipation through the rod.

6.2.3.4 Reproducibility of Process

Material accumulations generated by laser melting are intended to be used, among others, as preforms for bulk forming processes. The shaft may enable simplified handling and fixture if necessary. It is also possible to use the shaft as a functional component, for example as an axle when forming the preform into a gear wheel.

Forming operations with closed forging dies require preforms with a very accurately defined volume [10] so that the resulting parts do not suffer any defects, such as incomplete form filling. Investigations on the laser-based free form heading process show that a sufficient accuracy for a subsequent forming operation is reached. Table 6.2 illustrates both the absolute and relative standard deviations in diameter determined in experiments on preforms with a shaft diameter $d_0 = 0.5$ mm.

6.2.3.5 Influence of Shielding Gas

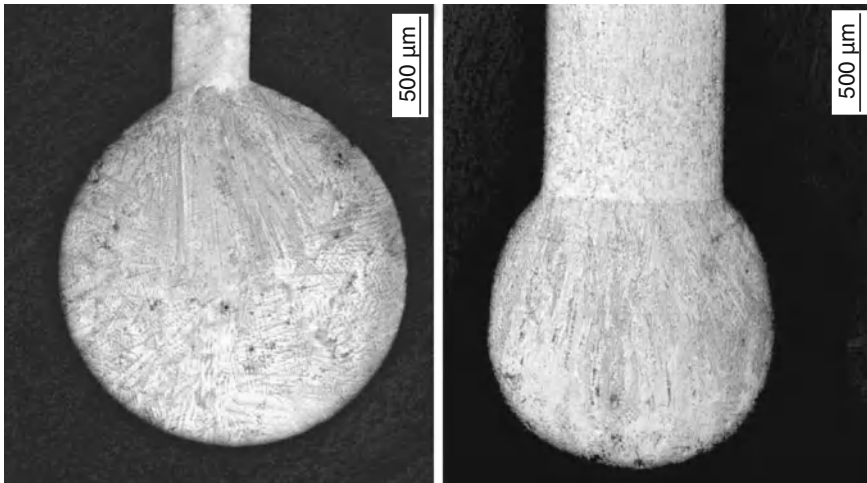
Although the adiabatic model neglecting heat dissipation is a proper approach, heat dissipation has a clear influence on the process and the result of thermal free form heading. With decreasing rod diameter d_0 , the relative amount of energy being dissipated by heat conduction also decreases, as well while the fraction of energy dissipated through the inert gas increases. Due to size effects, the surface/volume ratio increases with decreasing sample size. This leads to a rising

Table 6.2 Absolute and relative standard deviations of head diameter of preforms

P_L (W)	t_L (ms)	d_S (mm)	Standard deviations (mm)	Standard deviations (%)
64	500	1.136	0.004	0.39
64	1,500	1.598	0.005	0.34
64	4,000	2.135	0.010	0.49

importance in the characteristic inert gas parameters. Helium for example has, compared to Argon, a much higher heat transfer coefficient thus resulting in a different cooling behavior of the material accumulation. The cooling behavior of rather small material accumulations and small upset ratios u are mainly affected by the heat conduction within the rod and, directly at the end of the accumulation process, heat transfer due to radiation. So the heat inside the accumulation is dissipated through the rod so that the melt front moves from the beginning of the heat affected zone towards the top of the material accumulation. This can be seen in Fig. 6.19 in the orientation of the dendrites. The larger the upset ratio u the bigger the influence of the shielding gas and heat radiation. A not negligible part of heat is then dissipated by the inert gas. This at first leads to material solidification on the surface while the core is still liquid, resulting in the radial symmetrical orientation of the dendrites.

The time t_Σ , which is called the cycle time, is the duration until a material accumulation is generated by laser melting. It can be divided into irradiation time t_L and cooling time t_c . The irradiation time, which is equivalent to pulse duration, is directly linearly dependent on the aspired volume V_S of the material



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Fig. 6.19 Material accumulations; *left*: $d_0 = 0.5$ mm, $d_S = 2.25$ mm, $P_L = 80$ W, $t_L = 4,000$ ms. Radial symmetrical orientation of dendrites. *Right*: $d_0 = 1.0$ mm, $d_S = 1.63$ mm, $P_L = 139$ W, $t_L = 1,000$ ms. Orientation of dendrites coaxial to shaft

accumulation. As stated above, the part of the laser beam energy which is dissipated into heat for melting the rod varies, contingent on (among others) the irradiated material and laser beam wave length. Experiments have been carried out using a fiber laser with $P_{Lmax} = 300$ W, so that based on $d_0 = 1.0$ mm irradiation times t_L of a few 10 ms up to 10 s have been applied.

6.2.4 Formability of Preforms Generated by Thermal Free form Heading

The formability of preforms is not expected to be very good due the fact that a dendritic microstructure generally does not allow extended formability [23]. Experiments, similar to compression tests, have been carried out using a pneumatic forming machine and forming tools with coplanar, polished surfaces. The shaft of the preform is clamped by the lower die so that an axial movement of the preform is avoided. Preforms with different head diameters d_s have been formed with progressively increasing punch forces, ranging from $F_p = 0.3$ to 16.0 kN thus leading a decrease in final head heights after the forming process. After unloading of the preform, the flattened surface areas as well as the head heights were measured so that the average natural strain φ of the head can be determined. Figure 6.20 shows the dependence of the yield stress k_f on the average natural strain φ . It can be seen that the yield stress increases linearly with decreasing value of average natural strain (valid for $\varphi < 0$) [5]. This behavior is generally expected, however, beyond this it is worth noting that there is a very good accordance in yield stress between the three different sizes of preforms, so that within this variety of head diameters no size effects concerning formability could be determined.

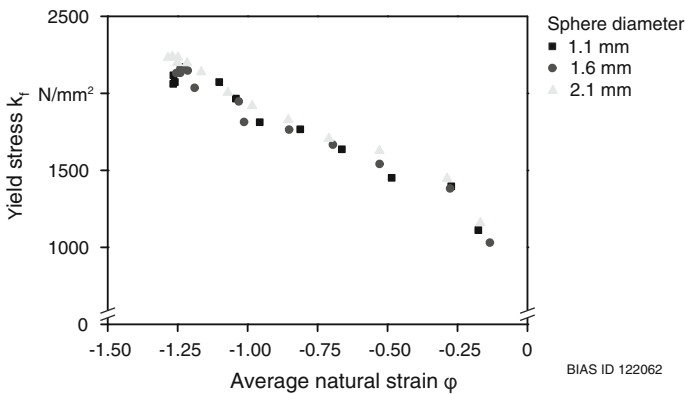
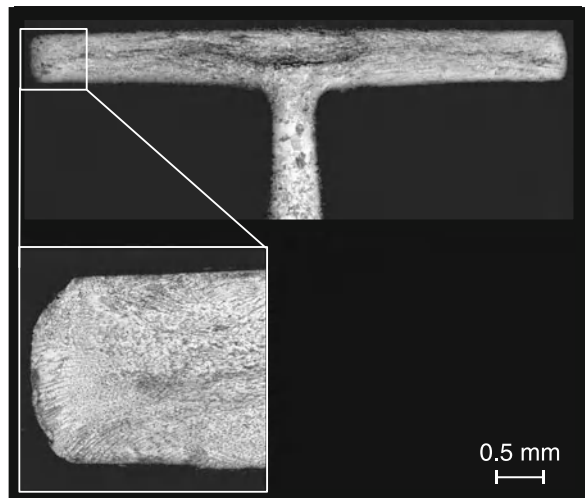


Fig. 6.20 Yield stress k_f dependence on average natural strain φ for different head diameters $d_s = 1.1, 1.6, 2.1$ mm

Fig. 6.21 Cross-sectional view of deformed preform $\varphi_S = -1.7$



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The forming limit of preforms is reached as soon as cracks appear. This is most likely to happen at the fillet between the circumference and the flattened head, as this is a strongly deformed area [12]. Due to the fact that the forming tools used are of 42CrMo4 with a hardness of 55HRC, higher punch forces would have led to the plastic deformation of the tool so that the maximum formability has not been identified. Figure 6.21 shows a cross-sectional polish of a formed preform with $\varphi_S = -1.7$ as well as an enlarged section of a strongly deformed area where no cracks can be detected. The direction of the dendrites, whose orientation is a significant indication of the direction of the plastic flow of the material, clearly conforms to the behavior stated in [12].

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Chapter 7

Tool Making

Ekkard Brinksmeier

Latin

a	Distance (mm)
$A_{a, xy}$	Laser affected area (m^2)
a_{ec}	Width of cut (μm)
a_{la}	Laser track distance (μm)
AM_{us}	Ultrasonic amplitude (μm)
a_{pc}	Depth of cut (μm)
c_p	Specific heat capacity ($J\ kg^{-1}\ K^{-1}$)
c_R	Material optical reflectivity
c_α	Optical absorption coefficient (m^{-1})
d	Diameter (μm)
D	Diffusion coefficient
d_G	Diamond grain size (μm)
d_{la}	Laser beam diameter (μm)
d_{min}	Laser subtracted diameter (M)
E	Elastic modulus (N/mm^2)
EA	Activation energy (J)
F	Force (N)
F_N	Normal force (N)
f	Focal distance (m)
f_{la}	Pulse frequency (laser repetition time) (Hz)
f_{us}	Ultrasonic frequency (Hz)
f_z	Feed per tooth (μm)
HK	Knoop hardness
h_a	Laser affected depth (m)
h_c	Chip thickness (μm)
h_{cu}	Uncut chip thickness (μm)

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$h_{cu,crit}$	Material specific, critical uncut chip thickness (μm)
$h_{cu,max}$	Maximum uncut chip thickness (μm)
$h_{cu,min}$	Minimum uncut chip thickness (μm)
h_s	Laser subtracted depth (m)
h_λ	Optical penetration depth (m)
I	Incident intensity (W m^{-2})
I_0	Peak incident intensity (W m^{-2})
I_{ab}	Peak absorbed intensity (W m^{-2})
i_{ch}	Chemical activity
j	Number of cutting edges
K	Thermal conductivity ($\text{W m}^{-1} \text{K}^{-1}$)
K_c	Fracture toughness ($\text{MPa} \cdot \text{m}^{1/2}$)
K_F	Faraday constant
K_p	Preston constant
k_λ	Optical damping constant
p	Pressure (N/mm^2)
P_{la}	Average laser power (W)
PR_R	Reactive sticking probability
P_{us}	Ultrasonic power (W)
Q'_w	Specific material removal rate ($\text{mm}^3/(\text{mm} \cdot \text{s})$)
Q_w	Material removal rate (mm^3/s)
r	Radius (mm)
r_0	Radius of the diffusing particles
R_a	Roughness, profile, arithmetic (nm)
R_z	Roughness, profile, average maximum height (m)
$r_a, x(y)$	Laser affected radius (m)
r_L	Laser beam radius at the optical lens (m)
r_M	Laser melted radius (m)
$r_s, x(y)$	Laser subtracted radius (m)
r_{th}	Thermal penetration radius (heat affected zone) (m)
r_w	Laser beam waist radius (m)
r_β	Cutting edge radius (μm)
S_a	Roughness, area, arithmetic (nm)
S_k	Roughness, area, core roughness depth (nm)
S_{pk}	Roughness, area, reduced summit height (nm)
S_q	Roughness, area, root mean squared (nm)
S_{vk}	Roughness, area, reduced valley depth (nm)
t	Time (s)
T	Absolute temperature (K)
T_0	Ambient (initial) temperature (K)
T_B	Absolute temperature of boiling (K)
T_M	Absolute temperature of melting (K)
T_{sc}	Absolute surface temperature (K)
t_B	Material boiling time (s)

t_D	Dwell time of laser-material interaction (pulse duration) (s)
t_{e-ph}	Electron-phonon scattering time (s)
t_R	Material removal time (s)
$\dot{V}$	Volume material removal rate (m^3/s)
$\dot{V}_e$	Flow rate of the etchant (ml/min)
v_c	Cutting speed (m/min)
v_f	Feed velocity (mm/min)
v_f	Feed rate ($m\ s^{-1}$)
v_{rel}	Relative velocity (m/s)
v_{scan}	Scanning speed ($m\ s^{-1}$)
v_t	Tangential velocity (mm/s)
v_z	Drilling velocity ($m\ s^{-1}$)
Z	Densification ratio
z_n	Powder bed height for the nth layer (mm)
z_T	Platform lowering distance (mm)

Greek

α	Thermal expansion coefficient (K^{-1})
β	Angle of incidence ($^\circ$)
γ	Rake angle ($^\circ$)
δ	Thickness of the Nernst layer
ΔH_M	Enthalpy of melting ($J\ kg^{-1}$)
ΔH_V	Enthalpy of vaporization ($J\ kg^{-1}$)
ζ	Beam quality factor
η	Dynamic viscosity of the liquid
θ_g	Angle of grit blasting ($^\circ$)
θ_{mf}	Angle of molten material front ($^\circ$)
θ_0	External angle of laser light incidence ($^\circ$)
ρ	Material density ($kg\ m^{-3}$)
ρ_{bulk}	Density of the consolidated powder (g/cm^3)
ρ_{powder}	Density of the loose powder (g/cm^3)
κ	Thermal diffusion coefficient ($m^2\ s^{-1}$)
κ_{el}	Electron diffusion coefficient ($m^2\ s^{-1}$)
λ	Optical wavelength (M)
ϕ	Laser fluence ($J\ m^{-2}$)
ϕ_{ab}	Absorbed laser fluence ($J\ m^{-2}$)
ϕ_{Loss}	Fluence losses ($J\ m^{-2}$)
ψ_{el}	Electrochemical potential

7.1 Tool Materials

Alwin Schulz

Tool materials for forming processes require good ductility with high hardness and high wear resistance. In the case of micro cold forming, additional aspects have to be considered in the tool material selection. The tool shape requires structures of microscopic dimensions, i.e. edge radii, bores and notches in submillimeter sizes, which need to be formed or machined. Viscous lubricants should be avoided, because they hinder the further handling of the workpieces. The use of tools without lubricant accelerates wear and might aggravate corrosion. Wear and corrosion debris on the tool surface are unacceptable for working with microscopic structures. Thermal effects of friction require additional attention to chemical processes at the surface.

There are many material candidates for making micro forming tools. Monocrystals like diamonds or sapphire are very hard (10,000 HV, 2,300 HV) and can withstand chemical attack. However they are expensive and it is very difficult to make complex micro tool shapes with them. Ceramic materials, e.g. alumina, zirconia, silicon nitride, etc., are also potential candidates because of their high hardness and chemical resistance. In general such ceramics are produced by sintering or hot isostatic pressing, both requiring great efforts on giving the bulk ceramic full density. For micro tools extremely fine powders are needed as the base material for these ceramics to deliver fine structures. Handling of these powders requires great efforts in every process step (i.e. powder production, storage, transport, green body forming, etc.). Another challenge to achieve precise tool shapes is the control of shrinkage of the ceramic during consolidation.

Between ceramics and metallic alloys, cemented carbides are another group of interest for micro forming tools. In these materials fine carbides, usually tungsten carbides or titanium carbides, are embedded in a cobalt, nickel or iron metallic binder matrix. With fractions of the metallic binder down to 3 % and carbide sizes which can be less than 0.5 μ , cemented carbides are very hard (up to 1,650 HV), corrosion resistant and can show high toughness (e.g. WC6Co has a bending strength of 1,700 MPa and ultra fine grain of 3,600 MPa). The high stiffness of diamonds, ceramics and cemented carbides is another advantage of these materials. The Young's modulus of diamonds is about 1,200 GPa, and fine grained cemented carbides can achieve more than 650 GPa which is more than three times the modulus of elasticity of steels. On the other hand, the manufacturing route for cemented carbides is expensive. In particular, the synthesizing of ultrafine carbide powder is very complex. Furthermore, consolidation via hot isostatic pressing is also limited to simple geometries. Machining of these materials is mainly possible by grinding or electrical erosion. Among carbides only a few types [B₄C (3,700 HV), SiC (3,500 HV) and TiC (3,200 HV)] possess a very high hardness, but they are not usually used for increasing the hardness of tool steels. Nevertheless, TiC is frequently used in cemented carbides and especially in

Table 7.1 Types of hard materials and basic properties

Chemical binding	Examples	Toughness	Adhesion to metals	Hardness	Stability $-\Delta G$	Chemically inert
Metallic	Borides, nitrides, carbides of transition metals (e.g. TiB_2 , TiC , TiN , VC , WC , CrC)	***	***	**	**	*
Covalent	Borides, nitrides, carbides of Al, Si, B (e.g. B_4C , SiC , BN), diamond	**	*	***	*	**
Ionic	Oxides (e.g. Al_2O_3 , ZrO_2 , TiO_2)	*	**	*	***	***

*** high, ** medium, * low

Ferro-Titanites (Trademark of Deutsche Edelstahlwerke GmbH), where their stability during heat treatment of their steel-based binder phase is utilized.

The bonding strength of the hard materials to the binder phase is another aspect which has to be considered. While it is desirable within the tool material to ensure high fracture strength, a low binding energy is required by the workpiece to reduce adhesive wear.

Hard phases can be subdivided with respect to their chemical binding, which strongly influences the behavior (Table 7.1).

From Table 7.1 one can conclude that hard phases with metallic bonding show a good compromise as long as chemical stability is not the major requirement. Among the transition metal carbides, TiC has the highest stability and hardness but the lowest toughness, while Cr - and W -carbides behave the other way round. Vanadium carbide and niobium carbide are in-between.

Further potential candidates for micro forming tool materials are metallic alloys. Among them iron-, nickel- and cobalt-based alloys are widely used. For tool steels the hardness and the wear resistance strongly depend on the type of carbides that precipitate during solidification and heat treatment. Table 7.2 gives an overview of carbide types in tool steels, including the composition and the

Table 7.2 Carbide types in tool steels [54]

Type	Hardness	Structure	Chemical composition in weight %		
	HV		C	Fe	Alloying elements
M_3C	900	Orthorhombic	6.8	Rich (40–70 %)	$Cr (\leq 18 \%)$, Mn
$M_{23}C_6$	1,200–1,600	Cubic (complex)	5.5		$Cr (20–60 \%)$, Mo , W
M_7C_3	1,500	Hexagonal	9		$Cr (30–65 \%)$, Mo , W , V
M_6C	1,500	FCC	2		W , Mo , $V (\Sigma = 60 \%)$
M_2C	1,800	Hexagonal	6	Poor (2–6 %)	W , Mo , $V (\Sigma = 80 \%)$
MC	2,800–3,000	FCC	16		$V (= 40 \%)$, W , Mo

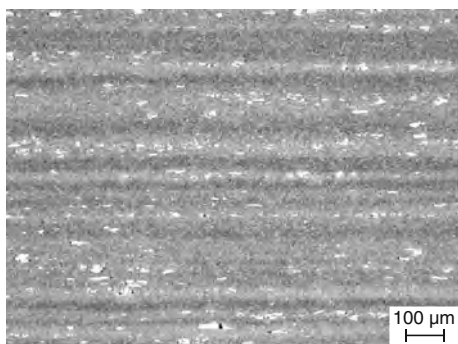
hardness of the carbides, with tungsten carbides and monocarbides from vanadium and tungsten having the highest values.

A major challenge for preparing metallic alloys is their tendency to segregate during solidification and the presence of large inclusions, when the melt is cast. An example is hypereutectoid 8 % chromium tool steel, which is produced by electro slag remelting (ESR), a steelmaking process which already delivers very high cleanliness and a lower degree of segregation compared to other melting and casting methods. The longitudinal section of a forged bar shows segregation bands with distances in the range of 100 μm and with some large chromium carbides (Fig. 7.1). If the requested geometry of the tool is of similar dimensions it might interfere with these material's structures and problems might occur either during machining or in service. If such alloys are to be used for micro forming tools, the microstructure of the alloys needs further refining. On the other hand, in particular for iron based alloys, the properties of the materials can be altered by specific heat treatments, giving excellent options for soft and hard machining according to cost effectiveness and a variety of design and manufacturing solutions.

Generally tools need to resist specific loads (e.g. wear, pressure, corrosion) in different areas. The selection of appropriate materials has to be aimed at the defined adaptation to the specific working behavior as well as to the reaching of an adequate service time for the tool. Solutions are provided by material surface modifications of a semi-finished tool or by direct combination of materials to a well defined compound with the desired properties.

Surface modifications have been produced for decades by plating, welding, thermal spraying, electroplating and physical or chemical vapor deposition, to name the most popular processes. These methods have in common that they produce a relatively sharp transition between the different materials in the micrometer scale or submicrometer scale. At the interface the binding is weakened since the chemical bond is limited due to different crystal structures and lattice parameters [109]. Furthermore, process-induced residual stresses show maximum gradients at such interfaces. Different thermal expansion of the materials can lead to additional stresses during in service heating, e.g. due to unlubricated friction contact [51]. Therefore these composites may only exhibit limited stability [97].

Fig. 7.1 Microstructure of hypereutectoid 8 % Cr tool steel, produced by electro slag remelting and forging (longitudinal section of forged material)



A countermeasure to enhance robustness is to generate a gradual transition of the material properties between the different functional materials. On coated surfaces intermediate layers are inserted between the materials with basic functionality which serve to mediate the differences in crystal structure, lattice parameters and thermal stresses [68]. In the case of a reactive component being involved in the coating process, e.g. in PVD or CVD processes (see Sect. 7.5), gradient layers can be developed just by the controlled variation of the reactive gas. Also multilayer components are realized, so PVD and CVD are the most promising techniques for generating tool materials with a gradient in chemical composition, not only of light elements (C, B, N) but also in terms of the dimensions of the micrometer scale.

Thicker coatings or bulks with layered graded structures scaled in the lower millimeter range can be developed during thermal spraying (see Sect. 7.1.2).

Another method of the modification of the surface material is to generate chemical gradients into the near-surface region produced by thermochemical processes. Here one component diffuses with a high diffusion rate into the base material, thereby forming a gradient (carburizing, nitriding, boriding, etc.) usually with the highest concentration of diffusing light elements at the surface. In all cases of surface alteration by coating or thermochemical processes, subsequent machining is limited to an adjustment of the surface, e.g. by grinding or polishing.

The preparation of tool composites of Fe-based materials is found in powder metallurgy, just by the HIPing of a steel powder with high amount of carbides with a core of a lower alloyed steel [100]. By powder mixing and adding of the powder layer by layer to a green body for sintering, one can achieve graded high-speed steel alloys, e.g. with significant variation in tungsten carbide [23]. But this procedure is limited by the powder particle size to only a few different powder layers per mm of gradient thickness. Several variants of the method for the manufacture of graded green bodies for powder metallurgy and technical ceramics were developed for this purpose, focusing on the handling of ultrafine powders to achieve thinner gradient layers [52]. Because of the small particle size of the powder required and a significant shrinkage during the hot isostatic pressing (HIP) of powders the techniques only show limited success, especially in the generation of semi-finished graded structures in the submillimeter range.

Spray forming, as a combination of melt atomization and subsequent direct consolidation, can be a more promising method to produce complex shaped gradients. Beam-assisted methods, e.g. laser beam melting, where the powder is deposited in thin layers and consolidated layer by layer, are also promising techniques (Sect. 7.2). In the following sections four technologies for micro forming tool materials production including gradients are presented: spray forming, thermal spraying, laser beam based sintering and selective laser beam melting.

7.1.1 *Spray Forming*

Spray forming is a primary forming technique to achieve fine grained materials from a melt with a high production rate. This technique is located between ingot casting and powder metallurgy with respect to production speed, fine microstructure and the isotropy of properties. Spray forming was primarily developed for the speeding up of powder metallurgical consolidation processes, like sintering or hot isostatic pressing. The technique is able to produce 4 t of high-speed steel billets in less than 2 h [91].

The main advantages of this technique are observed when working with alloys with eutectic phases and hard and brittle phases that directly precipitate from the melt. For high alloyed steels the former are mainly chromium-containing cold work alloys with eutectic chromium carbides, the latter are high-speed steels with high amounts of monocarbides (MC, mainly tungsten carbide and vanadium carbide). Figure 7.2 demonstrates this feature of the refining and homogenization of the microstructure by the spray forming technique (compare with Fig. 7.1 for the same steel).

In spray forming, the alloy melt is prepared by conventional ladle metallurgical measures. Usually the melt is kept in a vessel with a protective atmosphere and no slag is used either for protection or cleaning from residual impurities (Fig. 7.3). The melt with a superheat of 100–200 K is then poured into a funnel-shaped tundish. Ceramic foam filters at the bottom of the tundish may be used for cleaning the melt from oxides and other high melting point impurities. When leaving the tundish as a free-falling stream the melt is atomized to a spray cone by a concentric arrangement of gas nozzles. This spray cone consists of droplets with sizes in the range of a few microns to a few hundred microns, depending on the alloy type, diameter of the tundish opening, atomizing gas type and pressure. Furthermore, these droplets are accelerated by the gas flow to speeds of about 100 m/s. In contrast to powder metallurgical production, where the atomized melt is collected solidified as a powder, in the spray forming process a rotating disk is positioned in the spray chamber. It works as a substrate to collect the droplets before complete

Fig. 7.2 Microstructure (longitudinal section) of spray formed hypereutectoid 8 % Cr tool steel after forging, hardening and tempering (same alloy as in Fig. 7.1)

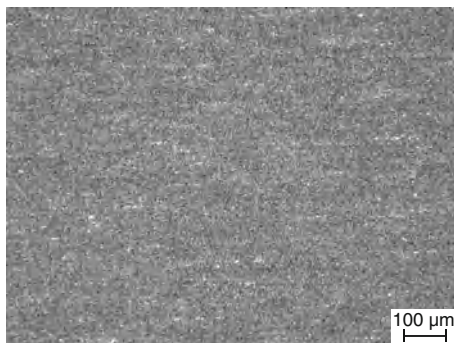
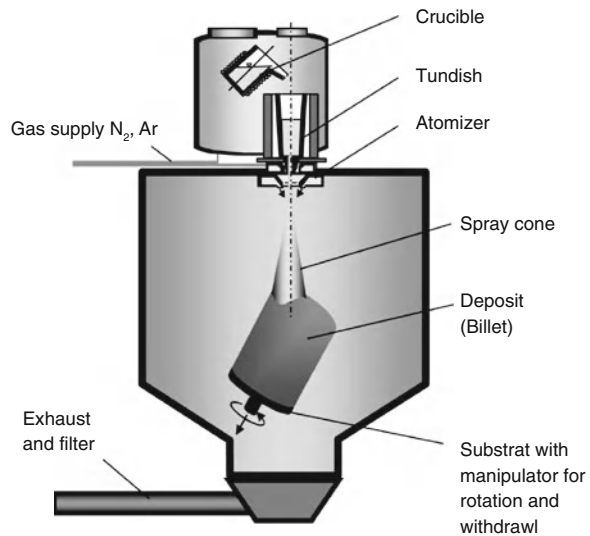


Fig. 7.3 Schematic of spray forming of a cylindrical billet



solidification and a billet-shaped deposit is formed while the disk is withdrawn in the direction of the rotation axis.

After atomization there is a high amount of enthalpy dissipated from the spray cone and the fine droplets are completely solidified. Large droplets on the other hand are still liquid but may be undercooled. There are also fractions of droplets in the spray cone in a semi-solid state. Upon impinging the surface of the substrate, the liquid droplets start to crystallize while the fine solid particles start to absorb heat from the larger droplets and are completely or partially dissolved. Therefore a thin surface layer, called a mushy layer, is created where the alloy is partially solidified and subsequently completely solidifies while the mushy layer continuously moves in the direction of the deposit growth. In addition, the momentum of the impinging droplets is transferred to the mushy layer, leading to disturbances in the liquid and fragmentation of the solidified structures, e.g. dendrite arms or acicular precipitates.

Finally, a fine and equiaxed microstructure results. Because of this sequential solidification from the mushy layer the microstructure only slightly depends on the position within the billet and the billet dimensions, as indicated by the eutectic cell diameters over the radius and for different sizes of high-speed steel billet (Fig. 7.4).

Due to sequential solidification during spray forming it is possible to form more complex deposit shapes. The use of a rotational tubular substrate leads to a tubular billet, which can either be used as a clad product [108] or removed. By moving a substrate plate through the spray cone and scanning it perpendicularly to the moving direction via the oscillation of the gas nozzles, a plate can be spray-formed. By using the even more complex movement of a negative substrate form it is possible to directly replicate a tool shape like a deep drawing tool in the automotive industry [71] or a stamping tool [111].

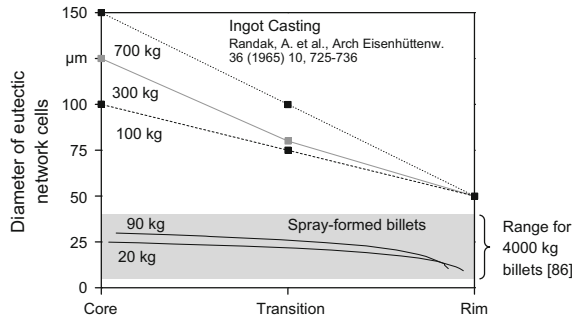


Fig. 7.4 Eutectic cell diameter over the radius and for different sizes of billet

Since the deposit is formed from disintegrated melt, i.e. droplets and particles, the occurrence of porosity, i.e. microvoids and pores, in the deposit is of particular importance. Porosity is classified into “cold porosity” and “hot porosity” according to its appearance and the mechanisms of its formation. Hot porosity indicates a hot spray cone with insufficient enthalpy dissipation. This leads to the entrapment of atomization gas in the compaction layer, which forms spherical gas pores (Fig. 7.5a). Gaseous dissolutions in the alloy melt (hydrogen, nitrogen) might also be expelled from the solidified alloy and recombine into similar looking pores (“Type A” porosity). The second feature of hot porosity (“Type B” porosity) are voids at the edges of dendritic cells due to the shrinkage of the residual melt (Fig. 7.5b). In large billets, a very hot spray might lead to hot cracking, e.g. thermal stresses, together with large amounts of residual melt forming large voids in the central billet area. Cold porosity on the other hand is caused by a lack of liquid material in the spray cone and in the mushy layer, where the initial fraction of liquid becomes too low, i.e. the compaction conditions are too cold. This leads to interstitial voids with a typical irregular shape (Fig. 7.5c), which frequently occur at the substrate or close to the surface of the deposit.

For large billet-shaped deposits, cold porosity at the rim can in many cases be neglected or, if necessary, can be removed by peeling or grinding prior to further processing. For tubular or flat deposits as well as for more complex shapes rim porosity is a significant problem. So the planning and running of a spray forming

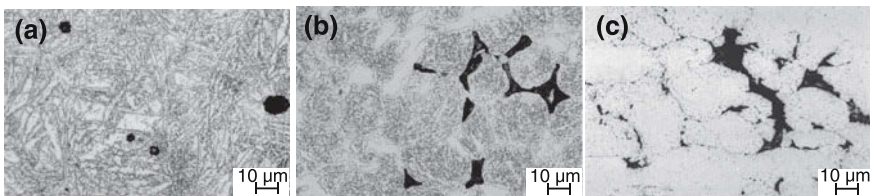


Fig. 7.5 Types of porosity in spray-formed steels. **a** Hot porosity “type A” (spherical gas pores). **b** Hot porosity “type B”. **c** Cold porosity (lack of liquid leading to interstitial voids with irregular shape)

process requires maximum efforts to overcome this problem. The heat transfer at the substrate has to be adapted by the choosing of a suitable substrate material. Pre-heating will help to reach thermally stable compacting conditions quickly and reduce cold porosity at the bottom. The spray forming parameters need to be properly adjusted to achieve an appropriate droplet size spectrum and appropriate fraction of liquid in the spray at the compacting position. The important parameters are the superheat of the melt, the metallostatic height in the tundish and the orifice at the tundish, which define the melt mass flow, the gas mass flow and the particle flight distance. A further influencing factor is the relative movement of substrate and spray-cone, which is first and foremost responsible for the shape of the deposit. The sum of movements is of utmost importance for keeping the thermal condition of the compaction layer at the right level. This leads to optimum compaction conditions concerning the cooling of surfaces outside the spray cone by the surrounding gas stream. Positioning of the spray cone scanning turning points outside the deposits can be used to smooth the mass and enthalpy input in complex deposit shapes. This measure of course reduces the yield. On complex-shaped deposits with high requirements of homogeneity in the microstructure, additional corrections by machining will be necessary.

Due to its specific features, spray forming allows the generation of new alloys or material combinations. In the field of micro metal forming, graded materials are of particular interest. Gradation of material composition can be done in several ways during spray forming.

At first, the composition of the alloy can be changed during pouring, either by controlled dissolving components in the tundish or by the mixing of two melts by simultaneously pouring into one tundish. As an example, Fig. 7.6 shows the radial

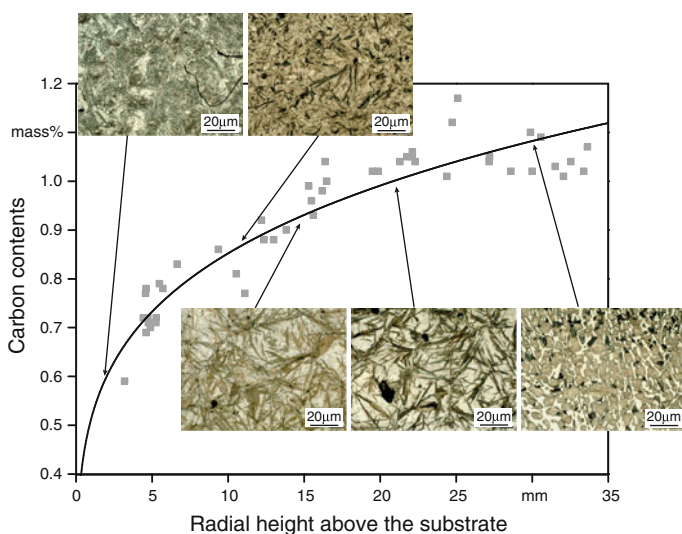


Fig. 7.6 Carbon gradient in low alloyed steel (spray formed tube, as-sprayed)

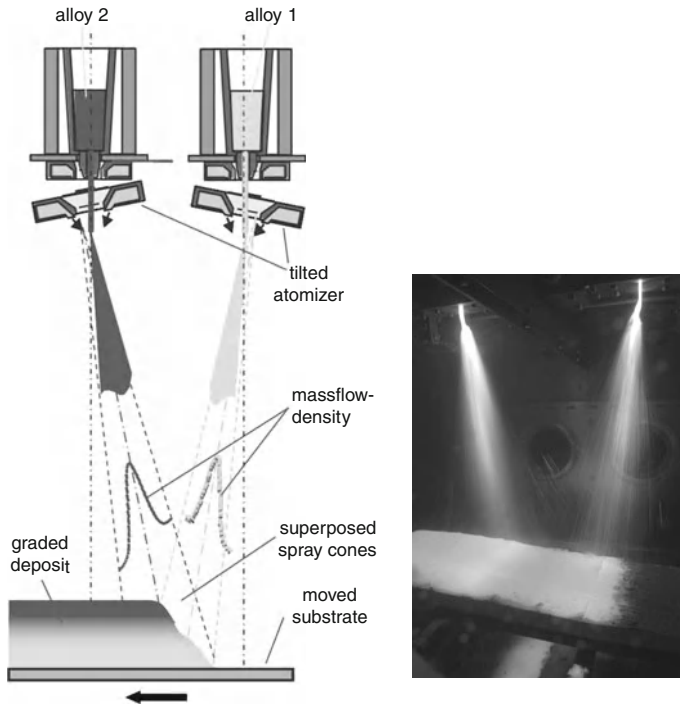


Fig. 7.7 Spray forming of a two-layered deposit with a graded zone

gradient of carbon in a tubular deposit [105]. This was achieved by continuously increasing the melt's carbon content by the further dissolution of carbon during the spray forming run.

Secondly, solid particles could be injected into the spray cone with increasing rate [18]. In this case a gradient will also be formed in the direction of deposit growth.

The third method uses two sprays of different melts from independent tundish and atomizers (Fig. 7.7). The different alloys, expressed by the two colors, superpose while forming the deposit. The tilting angle together with the nozzle-to-deposit distance affects the grade of the overlapping of the spray cones. If the axes of the two spray cones are parallel, two successive layers of two different material compositions will be deposited. At the interface there might only be slight transition zones because some elements will diffuse over a short distance at high temperature according to the difference in concentration between alloys.

Tilting the atomizers leads to the overlapping of the spray cones and a mixing of the droplets of the two different alloys before the complete solidification of the compaction layer. Due to turbulence in the semisolid surface layer the liquid fraction will continue mixing. In addition, it can be assumed that further homogenization of the composition of the alloy in the region of the gradient occurs by diffusion at liquid/solid or solid/solid alloy boundaries. This mechanism is well

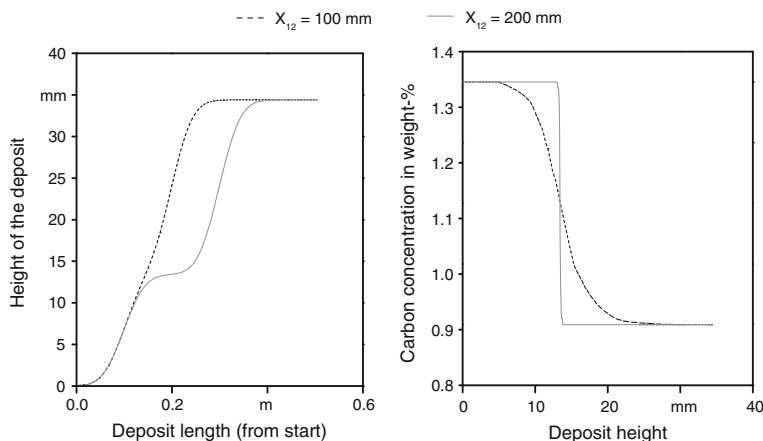


Fig. 7.8 Calculated results of deposit thickness (*left*) and vertical concentration profile (*right*) of a spray formed gradient deposit. Base alloys carbon concentrations are 0.9 and 1.3 wt%, respectively

known in powder metallurgy as “diffusion-alloying” but only works over short distances and primarily for light elements [85].

With the assumption of a Gaussian mass distribution within two spray cones and the scanning of the sprays perpendicular to the translational direction of the substrate, the height of the resulting flat deposit and the concentration profile in the deposit can be calculated. In Fig. 7.8 this is demonstrated for two alloys with different carbon contents, where X_{12} is the distance between the two spray cones. The concentration profile reveals a graded zone with a slope that strongly depends on X_{12} . In this way, together with post-heat treatment and in combination with various alloying elements, gradients of hardness, i.e. strength and wear resistance, can be achieved. The height of the gradient depends on the two metal mass flows, the mass distributions in the spray cones and their distance to each other, as well as on the spray distance and the translational speed of the substrate.

For spray formed billets the post-processing technique is very similar that of ingot castings. Usually they are hot-worked to close the residual porosity and to deform and recrystallize the grains to achieve a fine microstructure without segregation (compare Fig. 7.2). Eutectic carbides even in an as-sprayed state are elongated to some extent. Therefore during hot-working they align in the main forming direction. This causes some anisotropy of mechanical properties in the material. Sometimes even fracture of the eutectic carbides is observed and the matrix fills the gap between the fragments, thus refining the carbide distribution even more. While large billets are usually hammer-forged in a preliminary step, hot-rolling is the preferred method to form semi-finished long products. In general, because of the homogenous size and equiaxed shape of primary crystals and the small hard phases, deformability of spray-formed materials is higher in comparison to as-cast materials.

Another method to get rid of the residual porosity (closed porosity) of spray-formed alloys is hot isostatic pressing. This is the preferred method to densify spray-formed materials in more complex shapes. Because the amount of porosity of spray-formed materials, e.g. tool steels under optimized spray-form conditions, is less than 2 %, no encapsulation is necessary and the shrinkage problem is minimized. In principle these methods can also be used to reduce the width of the gradient in a plate or to form a curved graded layer, e.g. by bending or die forging.

The production route of high-alloyed tool steels does not influence the hardening and tempering behavior of the tool steel alloy [84]. For layered or gradient materials the situation is much more complicated because the heat treatment must be a compromise for all the alloys involved or a “gradient heat treatment” must be applied.

7.1.2 Thermal Spraying

Thermal spraying includes a group of processes in which a feedstock material (powder or wire) is molten by plasma, electric arc or combustion flame of liquid or gaseous fuel. The molten material is then accelerated by the expanding gases onto a surface, where it rapidly solidifies. The maximum particle speed mainly depends on the process and ranges from 40 m/s in flame spraying to 800 m/s in high velocity oxygen fuel spraying, with arc and plasma spraying in between and detonation gun spraying in the upper range. In the variant “cold spraying” the feedstock powder is highly accelerated (up to 1,200 m/s) but its temperature remains below solidus. The solid particles plastically deform and flow upon impact, creating hydrodynamic flow instabilities at the interface between the particles and the already deposited material. This results in bonding at the interface.

According to the technical standard EN 657:2005, one thermal spraying method is molten metal spraying, in which molten material is used as feedstock for spraying. This method is rarely used, but the binding link between thermal spraying to spray forming can be seen.

Thermal spraying usually provides thick coatings in the range from around 50 μm to about 1 mm. The deposited coating materials can be metals, alloys, ceramics, plastomers and composites.

The structure of the coating is dominated by highly deformed and rapidly solidified droplets which bond mechanically together and adhere to the substrate, respectively (Fig. 7.9). Therefore the substrate usually needs to be roughened, e.g. by sand blasting. Porosity in-between the droplets and embedded unmolten particles characterize the structure. In the case of an ambient process atmosphere, broken oxide shells are also embedded. At high particle impact speed, local fusion or diffusion welding are assumed to be additional bonding mechanisms. In general, the coating density and adhesion to the substrate increases with increasing particle velocity.

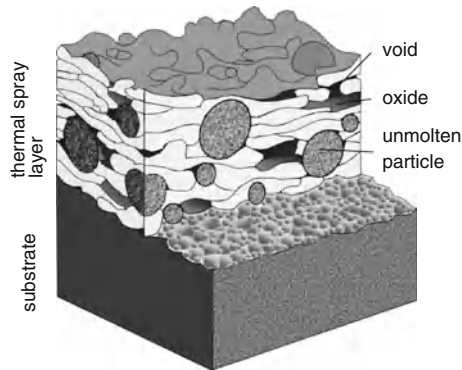


Fig. 7.9 Microstructure of a thermal spray coating (schematic)

An example of a thermal spray coating of chromium steel X20Cr13 (AISI 420) on mild steel shows the difference in microstructure of different thermal spray processes (Fig. 7.10). The rapid quenching of the droplets leads to a high hardness of the coatings: 590 HV0.1 for atmospheric plasma spraying (APS) and 705 HV0.2 for high velocity oxygen fuel spraying (HVOF). The hardness differences can be explained by the lower density in the APS coating (about 92 % of theoretical density).

Using plasma spray torches with stainless steel and 88 %WC-12 %Co powder, atmospheric plasma spraying can also be used for spray forming [113]. The quenching of thermal-sprayed droplets (HVOF) is so rapid that e.g. FeAl-alloys could be spray-formed to bulk materials with nanocrystalline structure [44]. Porosity in plasma spray coatings can be eliminated by hot isostatic pressing [92].

By using two thermal spray torches with different feedstock materials or by mixing the powders in one flame, the droplets can be mixed in the spray and simultaneously hit the surface. Variation of the two feedstock supplies during growth of the coating leads to a variation in the fractions of the two materials (e.g. NiCr-ZrO₂) in the direction of the growth of the coating, i.e. a gradient is

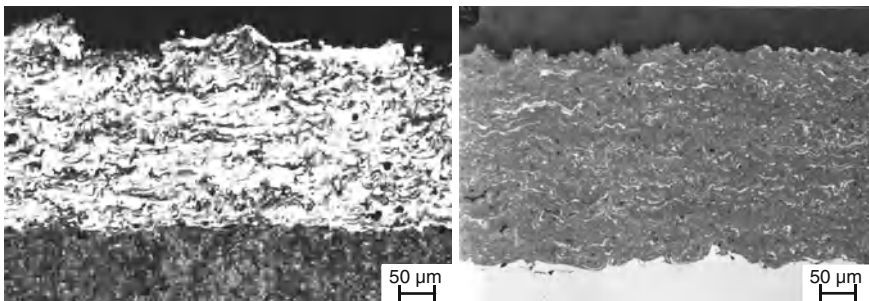


Fig. 7.10 Thermal spray coatings of steel X20Cr13 (*left* atmospheric plasma spraying, APS; *right* high velocity oxygen fuel spraying, HVOF)

generated [68]. The gradation is limited by the particle size since very fine powders are difficult to process. Furthermore, the thermal sprayed materials have a short thermal exposure and mixing of the different melt droplets or diffusion over larger distances is impossible. In fact in the “graded system” metal-matrix-composites or metal-mixtures with different mixing ratios are produced.

7.1.3 Properties of Graded Tool Materials

The properties of functionally combined materials are focused with regard to their specific applications and services. Micro forming tools demand the careful selection of materials or material combinations. Surface modification by coatings or diffusion layers, with more or less graded areas, can be advantageous. Post-heat treatments for desired microstructures and properties always influence the whole sample although it is possible to confine a heat treatment within the sample surface (e.g. short-term heat treatment by laser beam or high-frequency induction). Those treatments change the stress state of the whole sample. Under unfavorable conditions shear stresses at the interface of the combined materials might lead to material separation. In addition, combined materials usually have different coefficients of thermal expansion. Under thermal load this further changes the stress state and might promote cracking and/or delamination.

The resultant micro forming tool can reveal a very complex “property status” due to its elaborate design and manufacture. The orientation of graded or layered structures relative to the functional surface of the tool is also important, and the depth of loading also has to be considered (friction, abrasive wear, Hertzian stress). Finally, the surface roughness of the tool influences the forming process and the quality of the deformed workpiece, as it may cause tool failure due to stress concentration. To understand the performance of a tool made of a composite or graded material, it is therefore necessary to take the material features, the process characteristics as well as the specific loads in service into consideration.

7.1.3.1 Mechanical Properties

Important mechanical properties for cold forming tools are strength, toughness and fatigue resistance. Mechanical properties of a bulk material depend on the composition of the alloy and its microstructure. Experimental determination of these properties is more sophisticated when the samples, e.g. micro forming tools, become small. If the size of the compound becomes small, especially in the direction of the gradient and within the gradient, local properties may not be attainable from direct measurement. Here, and especially in the case of layered or graded material, finite element modeling (FE) is a promising method for property development and conditioning. On the other hand, one should focus on the relevant

material parameters of each material with respect to the specific loads. Additionally the determination of the interface properties is an important task.

In the last few decades there has been successful work to understand the behavior of such materials by FE calculations with mixing rules for the mechanical properties of the different phases (e.g. [76]) and with consideration of inner energy for crack propagation [28]. In modelling indentation of hard bodies into layered structures, even with significant abstractions to reduce computing time, e.g. the assumption of rotational symmetry, neglecting interfacial stresses, etc., the calculations showed good agreement with experimental results [45]. Nevertheless, indentation hardness measurements can also be used to estimate toughness on the edge cracks or coating delamination by the generation of complex 3-axis stress states below the indenter, showing that ductility of the layers needs to be considered and interfacial strength also needs to be introduced in these models.

Any kind of mechanical property can therefore be successfully modeled. Conversely, the desired mechanical properties are designable on demand via FE calculations, as long as the material can be produced in a specific way and the properties of the existing components and their interfacial behavior are well known.

As an example, the variation in impact toughness by slight variations in the constituents of high alloyed tool steels, spray formed (SF) and powder metallurgical (PM), which are both candidates for micro forming tool materials, are shown. It is dependent on the heat treatment, resulting in the corresponding hardness values and on the alloy composition (Fig. 7.11). This can be explained by the different amounts of hard carbides MC, M_2C , M_6C and M_7C in these alloys, which support crack propagation along the carbide/matrix interface in different ways. A softer matrix caused by a higher tempering temperature slows down crack propagation. Higher impact toughness in addition results from smaller carbides, i.e. the path of the crack along the carbide before reaching the matrix is shorter. This explains the difference between the SF alloys and the PM variants, and also the

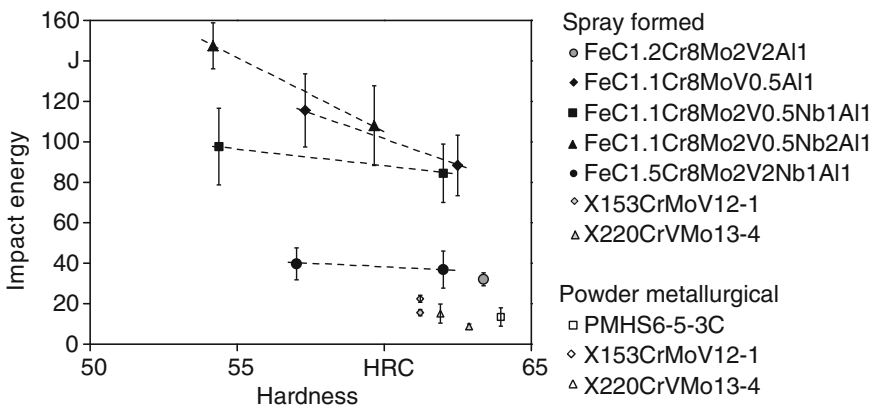


Fig. 7.11 Impact energy of high alloyed tool steels (sample $55 \times 10 \times 7$ according to “SEP 1314”) versus hardness due to tempering temperature (secondary hardening)

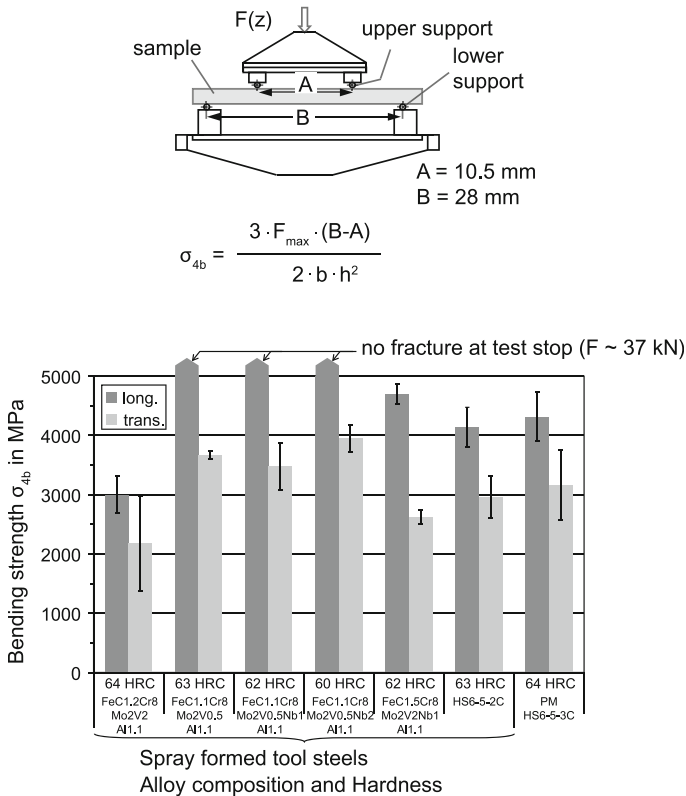


Fig. 7.12 Top sketch of the four-point bending set-up, Bottom four-point-bending strength σ_{4B} of different variants of tool steels (samples: $l = 35 \text{ mm}$, $b = 7 \text{ mm}$, $h = 5 \text{ mm}$) parallel and orthogonal to forging direction)

differences in alloys with different carbides. In the PM variants of high alloyed tool steels the average carbide diameter is 20–40 % smaller than in similar SF variants.

Beside resistance against impact, the mechanical strength of a tool steel is an important property. For tool steels with high hardness, which contain high amounts of carbides, it is hardly possible to perform tensile tests, because there is limited plastic deformation and the carbides must be considered as inner notches, thus leading to high statistical uncertainties. Therefore the bending strength σ_{4B} , which is determined by the four-point bending test, is a better measure to compare the static mechanical properties of those steels (Fig. 7.12 left). In this test the bending moment is constant between the rolls of the upper support and there are no high stress concentrations. This reduces the uncertainties for the test of samples with randomly distributed inner defects. From the permitted deflection of the sample in the z -direction ($z = 2 \text{ mm}$) in the sample mount the maximum stress in the set-up used is limited to about 5,000 MPa (i.e. $F \sim 37 \text{ kN}$).

The results show that there is a strong dependency of the bending strength on forging direction, which can be explained by the orientation of elongated eutectic carbides (Fig. 7.12 right). In the transversal direction, e.g. perpendicular to the longest Feret diameter of the carbide, there are a high number of short boundaries between the matrix and carbides. Therefore cracks can be easily initiated and propagate on a short pass vertically through the sample. In the longitudinal direction there are less carbides per length and therefore crack initiation is reduced. In addition, the path of cracks around a long side of a carbide is longer, e.g. more energy is needed to generate the new surface area, thus explaining the higher longitudinal values.

From this context it becomes evident that the amount of carbides and their morphology, both depending on the alloy's composition, also influence the bending strength. Eutectic chromium carbides have larger aspect ratios than niobium carbides or tungsten carbides. From the σ_{b4} values one can see that one measure could be the increase of the Nb contents and therefore suppress some of the Cr-carbide precipitation. On the other hand, after the addition of carbon, more carbides precipitate. This decreases σ_{b4} to some extent. Another aspect, which becomes visible in Fig. 7.12, is the influence of the process, i.e. the solidification conditions. Despite containing approximately 0.4 % more carbon and 1 % more vanadium in comparison with the spray-formed HS6-5-2C (AISI M2) the powder metallurgical PMHS6-5-3C (AISI M3:2) shows similar bending strength. This can be explained by the faster solidification rate during power production, i.e. there are more but much smaller carbides in the PM material in comparison with the SF material.

In spray formed materials, a small amount of the process gas nitrogen is entrapped. It can influence material properties. An increased nitrogen content in a vanadium containing hot-working steel (755 ppm instead of 340 ppm) decreases the material ductility at 550 °C by about 12 % [87]. This can be explained by higher thermal stability of V(C,N) in comparison to VC at the grain boundaries, which cannot be completely dissolved during heat treatment. Embrittlement due to an increased amount of nitrogen in the vanadium containing tool steels was also observed in bending-tests and impact tests at room temperature, as well as during thermal fatigue testing.

As shown before for thermal sprayed coatings (Sect. 7.1.2), hardness is not only a matter of constituents and heat treatment but also of porosity, which can be interpreted as an additional very soft phase with poor binding.

7.1.3.2 Wear Resistance

Wear in general occurs at the surface of a tool in contact with a medium or another solid material. Basic wear mechanisms are abrasive wear, adhesive wear and corrosive wear as well as surface fatigue (see also Sect. 2.2.2). They all have to be considered for tool wear in micro forming. In specially designed tools with a material gradient at the effective surface, the wear resistance may change locally.

Disturbances in micro forming can occur, including failure of the complete tool due to a local increase in friction and wear and therefore stresses. Such effects have to be considered during tool development. However, considering the responsible wear mechanisms, it may nonetheless be possible to use graded or layered functional surfaces. Also the final surface protection of graded materials can help achieve equivalent wear conditions.

In the case of dry friction, but also in mixed friction or with hard particles between the sliding materials, abrasive wear is the main wear mechanism. Hard materials under sliding contact at the sample surface, e.g. inclusions or already strain hardened areas in the workpiece, embedded or free particles, e.g. dust or corrosion debris, scratch the tool surface. This leads to local plastic deformation, e.g. debris from micro-cutting with material removal from the tools surface. With increasing tool hardness the wear is reduced. Homogenously distributed monocarbides or eutectic carbides prevent ploughing by hard particles. Furthermore, abrasive wear resistance depends on the carbide types. In particular monocarbides from tungsten, vanadium or niobium are much harder (NbC, WC about 2,500 HV0.05; VC up to 3,200 HV0.05) than $(\text{Fe}_{80\%}\text{Cr}_{20\%})_7\text{C}_3$ with 1,500 HV0.05 or 1,670 HV0.05 if they additionally contain 2 % vanadium, as compared in Fig. 7.13. The variation in carbon content (for example produced by spray forming for enhancement of tool strength) leads to different wear resistance. This can be considered in the tool design.

A second wear mechanism with major importance for micro forming tool applications is adhesion. In this case the tool and the workpiece are locally welded

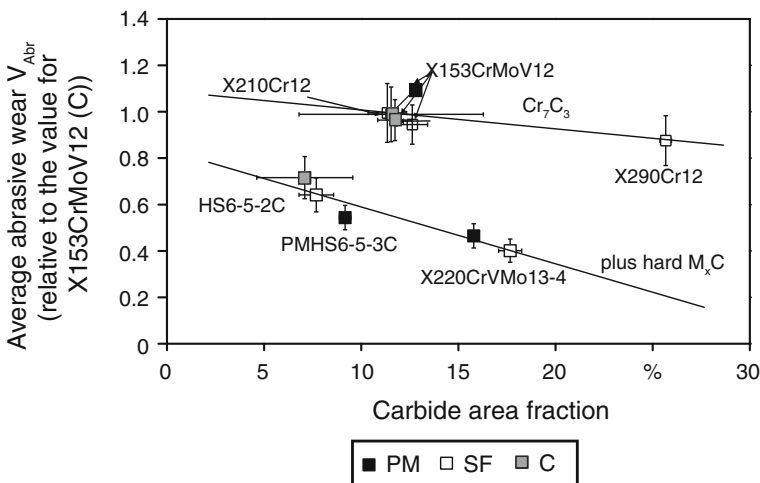
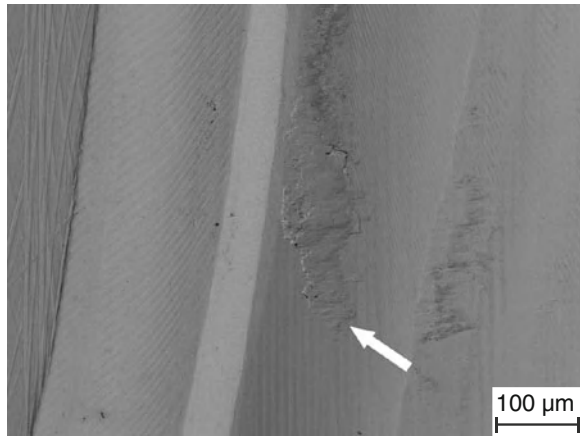


Fig. 7.13 Abrasive wear resistance (pin on-disk and rubber-wheel test according to ASTM G65.94) of hardened and tempered tool steels (61–64 HRC) versus carbide content determined by light microscopy (average of wear losses of both tests relative to the loss of ingot casted X153CrMoV12 (AISI D2) cold work steel), *C* ingot casting, *SF* spray formed, *PM* powder metallurgical [88]

Fig. 7.14 Debris from austenitic stainless steel (arrow) on a micro swaging tool (back scatter electron image)



and torn apart which leads to material transfer between the wear partners. A typical example for adhesive wear debris on a micro forming tool is shown in Fig. 7.14. One solution can be the use of finely dispersed tiny V(C,N)-particles which strongly reduce the tendency towards adhesion in cold-work steels. Due to the similarity of V(C,N) and Nb(C,N), diffusion alloyed high-Nb steels are expected to show similar performance [41].

Adhesive wear can also be avoided by selecting a “less similar” tool material with or by coating of the tool to reduce the tendency towards welding. Oxidation of the tool surface can also reduce adhesive wear. Therefore a final oxidizing heat treatment can be taken into account before using the tool to thicken the natural oxide layer, especially when chromium or aluminum are dissolved in the surface area.

As micro forming works with high local loads, one has to consider elastic deformation in the region close to the tool surface, either due to friction or Hertzian stress by local contact. There cyclic loading mainly leads to shear stresses and later to fatigue cracks directly at or closely below the surface. During a number of loading cycles the cracks grow and meet to form fatigue wear debris, i.e. small or larger detached parts of the tool surface. When the tool is protected against wear, by a coating or hard surface layer, the layer thickness must be selected with respect to a maximum load, because the maximum Hertzian stress should not be located at the interface or below. On the other hand, if the protective surface layer has a low friction coefficient, but has limited fatigue resistance, it can be advantageous to reduce the thickness of this layer and to locate the maximum shear stress below the gradient in a material that can withstand the alternating shear stresses without serious fatigue for long load cycles. Gradients which reduce the internal stresses at the interface between the different materials might further reduce crack susceptibility.

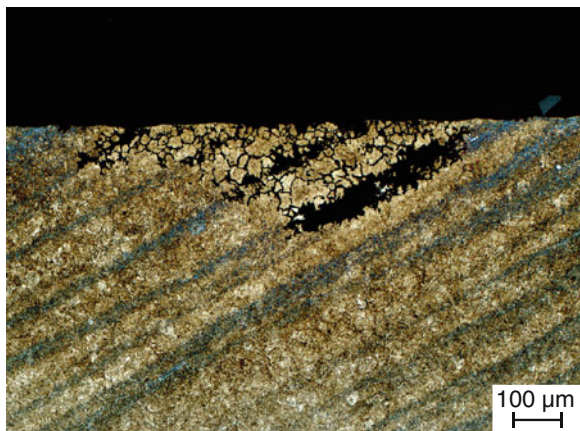
7.1.3.3 Corrosion Resistance

Corrosive attacks in principle occur at the surfaces of materials which are in direct contact with a corrosive medium that disintegrates the material dissolving it into the corrosive medium or by the formation of new, less stable compounds. For tool materials electrochemical corrosion is the most common type of corrosion. In this, a local Galvanic cell is formed a current flows and the less noble metal is electrolytically dissolved.

According to these mechanisms for metallic alloys in tool applications, the main reason for tool corrosion is segregation, precipitation and the mechanical electrical contact with other parts of the facility or with the workpiece material, which might form a galvanic couple. In dry cold forming, moisture and chloride from unprotected handling are the main sources for the electrolyte. Corrosion of the tool anyway leads to a loss of tool material, i.e. corrosive wear. Figure 7.15 shows an example of severe corrosion caused by abrasive particles in a corrosive medium (erosion corrosion). The corrosion is promoted by the segregation banding of the continuous cast and hot-worked steel. This example shows that more homogenous material, e.g. produced by spray forming, is expected to be less susceptible to selective attack by corrosion. Looking at the gradients or layered structures, these structures can be regarded in a similar way like segregations. As long as the structure is graded or layered parallel to the surface only the outer layer determines the corrosion behavior of the tool. In other cases where the layers are inclined relative to the surface the weakest material determines the corrosion behavior (Fig. 7.15).

The principal protection measure against corrosion is either the use of more noble metals or the use of protective coatings, which might be natural, e.g. alumina on aluminum, chromium oxide on stainless steel (steel with more than 10.5 % Cr in dissolution) or artificial, e.g. galvanic chromium coatings or ceramic coating. For corrosion resistant tools the main feature is the stability of the protective coating. In particular, the natural passivation layer on chromium steels forms

Fig. 7.15 Severe corrosion of a conventionally produced tool steel X46Cr13 (AISI 420)



within minutes and the growth rate and thickness of the layer increase with temperature. But the layer is only in the range of 10 nm thick and can therefore be easily damaged, especially by abrasive wear.

Another aspect of corrosion protective coatings is the density of the coating layer. Porosity in a thin coating will reduce the corrosion resistance, as was observed when comparing APS- and HVOF-coatings of around 150 μm thick duplex stainless steel with a porosity of 1.5 % and less than 0.2 % [113]. In cases where there is no possibility of avoiding porosity, the materials can be infiltrated by polymers to fill the pores and enhance their corrosion resistance.

7.2 Additive Manufacturing

Knut Partes

7.2.1 *Basic Principles and Prospects*

Additive layer manufacturing (ALM) is a generative layer-wise process to produce objects from 3D model data. A near net shape geometry is desired. ALM can be used for the production of prototypes (rapid prototyping), for the production of parts (rapid production) or for the production of tools (rapid tooling).

In order to fabricate an object from a 3D model the following steps are essential. The first step is the 3D geometry slicing, corresponding to the layer thickness. The thickness of the layers depends on the specific type of the process or the demands on the vertical resolution. After the slicing step, trajectories have to be defined which describe the movement of the energy source in every plane. In order to define the trajectories the spatial resolution of one plane has to be taken into account. This resolution is named lateral resolution. The lateral resolution depends on the ALM process itself. With the information of the plane geometries and the individual trajectories, a “build job” for the ALM machine has to be generated. The diversity of the possible methods to actually generate the object is high. An overview of the historical development and the diversity is given in [56]. It was concluded that only three methods are significant for the production of dense metal parts. For the fabrication of tools only dense metal parts have the potential properties of strength, fracture toughness and wear resistance. Hence in the following only these methods will be taken into account.

These three techniques are Selective Laser Melting (SLM), Direct Metal Deposition (DMD) and Electron Beam Melting (EBM). All these processes have in common is that the parts are made from powder. For the generation of micro forming tools the aspect of the lateral and vertical resolutions is very important. The generation of micro tools is in general possible with all these techniques.

In SLM, a powder bed is placed with a defined thickness. The process is commonly placed in a chamber flooded with an inert gas atmosphere in order to

avoid oxidation. The laser beam is commonly scanned with a 2D scanner over the surface. The laser beam is generally focused with an f-Theta lens system on the surface. This lens system keeps the focus in one plane instead of a curved plane (petzval membrane). Hence the focal position is supposed to be independent of the scanner position in terms of height. The laser beam interacts with the loose powder bed and increases the local temperature above the melting temperature. The powder particles completely melt and locally fuse together after solidification. By moving the laser beam the melt pool moves over the surface. After the melting process the powder bed is partially consolidated. Afterwards the powder bed is lowered and a new layer is placed on top and the process repeats. A generated part density of 100 % is desired. However, the parts being generated do have a porosity of about 0.5–1 %.

The layer thickness is basically limited by the powder particle size. The lowest limit of the powder particle size in commercial machines is about 10 μm . However layer thicknesses of 30–50 μm are more realistic. The lateral resolution is physically limited by the melt pool size and powder particle size. The melt pool size is basically a consequence of the process parameters and the size of the laser spot on the surface itself. In general it can be assumed that the size of the melt pool follows the size of the laser spot. The line energy also influences the melt pool size. It is obvious that the resolution is also limited by the increments of the scanner, but in general this is smaller than the laser spot diameter. Moreover the size of the laser spot is generally larger than the size of the particles. A state of the art laser today is about 50 μm in diameter according to the manufacturers of the SLM machines.

In DMD the laser beam generates a melt pool on a surface. The powder is delivered by a coaxial nozzle to the melt pool. The particles are commonly fed via a pneumatic powder feeder. This powder feeding technique is basically developed for thermal spray processes. Hence it is optimized for the particle sizes of this type of process. The process is similar to the laser cladding process. A melt pool is generated with a laser beam. The particles travel through the laser beam, being partially heated, and enter the melt pool. In the melt pool the particles melt and solidify. After solidification a clad track is formed. By putting clad tracks next to each other a plane can be covered. One essential disadvantage is that the clad tracks can only be placed on the material below. Hence, overhang structures can only be produced up to a certain angle. The nozzle geometry is coaxial in order to get the flexibility of being independent of the moving direction in one plane. In this type of process one layer is welded by moving the complete working head (or the object) in the x- and y-coordinates. After processing one plane the table is lowered or the welding head is increased in height and the next plane can be processed.

Moving speed and accelerations are limited due to the process principles of the machine kinematics. Moreover, based on the pneumatic powder feeding technology, the particle stream cannot instantaneously be switched on or off. Even if the powder feeder is fast enough to switch on and off, the hoses that connect the powder hopper with the working head can cause latencies in switching. Hence in practical conditions the laser is switched on and off. The powder stream is more or less continuously running. A lot of powder is lost due to misalignment or laser off

time. These powder particles can only rarely be recycled because some of the powder particles have traveled through the laser beam. This causes an oxidation of the particles. Experiments show that the processing result decreases greatly in quality if a part of the powder particles was previously oxidized. Presently this powder cannot be decontaminated. Moreover some machines are constructed inside inert gas flooded chambers like the SLM machines. Some of the particles that have traveled through the laser beam remain as overspray on the surface.

The vertical resolution is limited by the minimum track height of the process. This track height is in general somewhat as large as the particles themselves. Thicknesses as small as to 20 μm [110] have been reported. The lateral resolution is basically limited by the track width.

The power density plays an important role in the consolidation mechanism for SLM and DMD. If the power density exceeds a certain threshold, the material will start to vaporize. If vaporization occurs, the melt pool will start to be shifted apart from the center of highest power density due to the pressure wave. A keyhole (capillary filled with metal steam) will be formed. As a consequence the melting depth is strongly increased. This is called the deep penetration effect.

Below the critical threshold of the power density the consolidation mechanism is mainly driven by the heat conduction effect. The thermo physical property of the material is dependent on the state of the material. The material can either be in the powder state or in solid state. It can easily be seen that the heat conductivity of loose powder is much less than that of the bulk material. It has been shown in models by [82] that the melt pool geometry can be described by modeling the heat conduction relationships taking the different heat conductivities into account. From this it can be concluded that the position of the laser beam has an influence on the consolidation result. E.g. it has been shown in [17] that the consolidation of an overhang structure needs less energy because the local heat conduction losses are comparably low. This is caused by the fact that consolidation is locally performed on loose powder. However, currently the development of an online in situ laser power control based on a pyrometric coaxial heat measurement unit are under investigation. This could be an important development for SLM machines. Another effect of local heat accumulation is described by [17] in U-turn situations. This is caused by the surrounding weakly heat conducting powder bed and the acceleration of mirror tilting. An online laser power control might avoid this accumulation.

Beside the power density, the scan velocity is another important parameter. If the scan velocity increases, the laser power has to be increased as well in order complete melting the powder bed. However, depending on the laser beam geometry and the material, balling occurs above a certain threshold. The effect of balling has been shown in many publications [110]. This is an effect correlated with the Rayleigh theory of the free propagating jet of liquid. Lord Rayleigh discovered that every continuous jet of fluid tears apart due to surface tension effects after a certain path length of free travel. This effect can be seen in every water tap. If the scan velocity of the laser beam rises with the laser power, the melt pool length becomes longer. If the length becomes critical, a similar effect occurs on the melt pool. As a result of that the material solidifies as droplets on the

surface. This effect, called balling, should be avoided. It is currently a limiting effect of the buildup rate.

The EBM process is more or less similar to the laser process, but the energy source for the powder consolidation is an electron beam. The electron beam is scanned and focused by magnetic fields along the surface and generates a melt pool. The advantage of the EBM process is the high dynamic of the electron beam. The downside of this process is that the process needs a vacuum chamber. Some commercial EBM systems are presently available. However, the process does not seem as well developed as that of SLM technology.

Vertical resolution is basically limited by the particle size itself, as in the SLM process. The lateral resolution is limited by the particle size and the size of the melt pool. The melt pool size is coupled to the size of the electron beam spot at the surface and by the current. Theoretically, the electron beam spot can be smaller compared to the laser beam spot size.

In EBM the melt pool is generated by an electron beam. This beam is focused and deflected by magnetic field lenses. Due to the fact that the electromagnetic field lenses lack mass inertia, the deflection can occur much faster than compared to the laser beam. The laser beam deflection is limited by the scanner servo inertia. This has one particular advantage; the melt pool geometry can be modified by a fast scanning beam in terms of a broadening or elongation of the melt pool by scanning the electron beam parallel or perpendicular to the moving direction. Even though the electron beam does not have the diffraction limit of a laser beam, the spot size of on the surface is typically $10\text{ }\mu\text{m}$ [31] on flat surfaces, e.g. for welding applications.

All these three processes are able to produce dense metallic parts. Therefore the processes are possible candidates for the fabrication of tools. However the least resolution and the limitation in the design in the DMD process makes this process probably the last choice for most applications in the fabrication of micro tools. The highest potential is given by the SLM and EBM processes in order to fabricate near net shape micro geometries out of tool steel materials.

Different methods for the placing of the powder exist. In general the building platform in SLM and EBM processes is lowered in order to place the following layer. A common technology is the usage of a loader filled with powder (Fig. 7.16a). The loader is moved over the previous layer and feeds the powder with a rotating cylinder

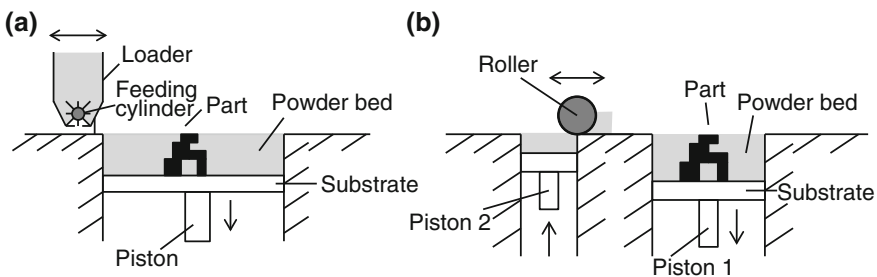


Fig. 7.16 Different powder placement mechanisms. **a** Loader. **b** Two piston strategy

drive. Afterwards a wiper makes the loose powder bed flat. Hence the height of the loose powder bed is given by the difference in height between the consolidated surface and the amount of height the previous layer was decreased. Another method of powder placing consists of two platforms (Fig. 7.16b). One cylinder is filled with the loose powder and the other platform contains the building platform. While the building platform is lowered the feeding platform is lifted. The changes in height represent the same amount of volume as the decreased layer. Hence the right amount of building platform and flattens the surface.

The SLM scanning speed is limited by the inertia of the galvanometer mirrors itself. One can see that the EBM process has the advantage of being a fast moving energy source. This allows generating nearly arbitrary shaped temperature distributions at the surface. However the surface finish and feature resolution of the EBM process is somehow weak in comparison to SLM. These aspects could be important for micro tool production. For high feature resolution SLM might be a better choice. If the powder material needs a specific heat treatment during the process, EBM could offer greater flexibility. All the features of EBM and SLM are listed in Table 7.3 according to [27].

There is no generally preferred production method for the fabrication of micro tools. However it can be seen that regarding the restriction in terms of resolution, the SLM process presently has the best performance. Moreover, the spectrum of treatable materials is slightly larger compared to EBM.

In order to process the powder particles in the EBM or SLM process the flux ability of the powder is important. In order to have a high resolution, generally small powder particles are needed. With a decrease in powder particle size, the ability to flow tends to decrease. This is caused by the comparably higher surface to volume ratio with decreasing particle size. If the surface has greater influence than the volume, adhesion forces between particles and hygroscopic effects makes the powder tend to agglomerate. Moreover, the lowest surface area can be achieved when the particles are spherical. However, depending on the atomization process at the powder fabrication, some particles can have satellites, some can be flat or they can have a very irregular shape. If the particles are irregular, clamping

Table 7.3 Comparison of features of EBM and SLM, [27]

Characteristics	Electron beam melting	Selective laser melting
Thermal source	Electron beam	Laser
Atmosphere	Vacuum	Inert gas
Scanning	Deflection coils	Galvanometers
Energy absorption	Conductivity-limited	Absorptivity-limited
Powder pre-heating	Use electron beam	Use infrared heaters
Scan speeds	Very fast, magnetically-driven	Limited by galvanometer inertia
Energy costs	Moderate	High
Surface finish	Moderate to poor	Excellent to moderate
Feature resolution	Moderate	Excellent
Materials	Metals (conductors)	Polymers, metals and ceramics

between the particles can occur. In order to avoid having water in the powder the powder should be placed in an oven before processing. Besides the oven, chemical treatment methods exist for drying the powder before use.

In addition to the average particle size, it has to be taken into account that the particle size distribution follows a logarithmic normal distribution from the manufacturing process. The average size and the distribution have an effect on the powder density. The powder density is always lower than the density of the bulk material. The ratio between the powder density and the density of the bulk material has an influence on the amount of porosity that has to be released during the melting process. This ratio is called densification ratio “ z ” in the following. Moreover, it has a direct consequence on the layer height after solidification. E.g. if the densification ratio is 0.4, the layer thickness will be reduced by the melting process down to 40 % of the height of the powder bed.

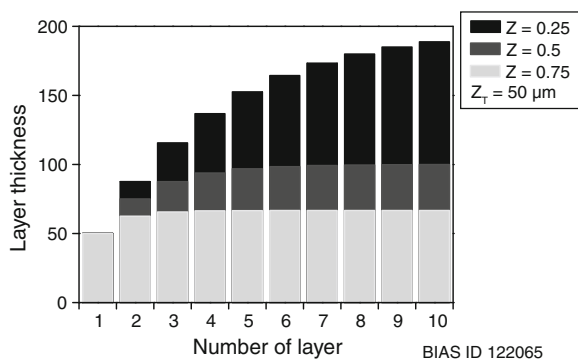
The step height result of lowering the building platform will remain the same in the complete process. The amount of powder thickness will increase from step to step due to the densification process.

The height of the first loose powder bed is z_T . After consolidation the density increases by the densification ratio Z . The thickness of the consolidated powder bed is now reduced to z_1 . The index “1” represents the first step. In the following step the platform is decreased by z_T . A loose powder bed is placed with the height of z_T plus the height that has been released in the previous step $1-Z$. This powder bed shrinks after consolidation to a height of z_2 . A general mathematical description of this phenomenon, called “asymptotically layer height”, is given below according to Meiners [65].

$$z_n = z_T \cdot \sum_{k=0}^{n-1} \left(1 - \frac{\rho_{\text{powder}}}{\rho_{\text{bulk}}} \right)^k \quad (7.1)$$

The resulting layer height of the n th layer can be calculated. By plotting this equation it can be seen that the final track height is reached after very few iterations. It can be seen that the limiting value depends on the densification ratio, Fig. 7.17. It can be concluded that z_n approximates z_T/Z for a large number of

Fig. 7.17 Asymptotic layer height development with various densification factors



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steps. Hence if the difference between the powder density and the bulk density is low, the change in the layer height over time will also be low.

In order to fabricate tools it might be necessary, especially for micro scales, to have geometries with a low amount of layers available. Moreover, the fabrication of tool steel places demands on the temperature time regime, hence the amount of energy density has to be chosen well. The aspect of asymptotic layer height can be important in these cases.

It can be seen that the change in the first layer is large. The energy might have to be adapted for critical geometries and/or materials for the first layers. Moreover, it can be concluded that the larger the densification ratio, the stronger the effect of the asymptotic layer height. Hence for structures with very few layers, a powder with a large Z should be chosen.

Different materials need different energies to melt. For SLM processes, different laser powers are necessary for different materials and/or geometric conditions. In order to achieve a high lateral resolution the spot size should be in the range of the particle's size. The minimum spot size is related to the beam quality of the laser beam, assuming that the optical configuration, e.g. the focal length of the f-theta system, is constant. Today, single mode lasers are available that supply laser beams very close to the physical diffraction limit. A high beam quality has the consequence of a high power density. As a consequence of the high power density, the energy losses in the lenses generate comparably high temperature gradients. The index of refraction is a function of the temperature. Hence the laser generated heat in the lens acts like a second lensing effect. Another consequence of the energy losses in the optics is the deformation of the surface caused by the thermal expansion coefficient. The sum of these two effects is known as thermal lensing.

The thermal lensing effect lowers the beam quality and hence increases the focus diameter. Moreover, the focus of the laser beam is shifted because of the change in the overall focal length of the lens. In general the focal length is reduced. This causes a movement of the focus away from the surface of the powder bed. Both effects, decreasing beam quality and focus shift, increase the overall spot size on the surface. This problem can currently be seen in some SLM systems using a single mode laser and conventional fused silica lenses.

Another important aspect of the thermal lensing effect is the temporal development of the thermal lens. This effect is called thermal blooming and is described in the following. Before the laser is switched on all the optics are cold and in equilibrium. After the laser is switched on, the optics heat up and the effect of thermal lensing becomes more and more dramatic. The timescale in which thermal blooming occurs is for a commercial system in the range of approximately one minute from no thermal lensing to the maximum effect (Fig. 7.18). This is also about the time used for the consolidation of one layer. The layer manufacturing time is strongly dependent on the size and amount of the parts that are fabricated. However the largest change in spot size occurs in the first minute. It now becomes very problematic, because the complete change of the spot size on the surface happens in one layer. While the next layer is placed the optics have time to cool

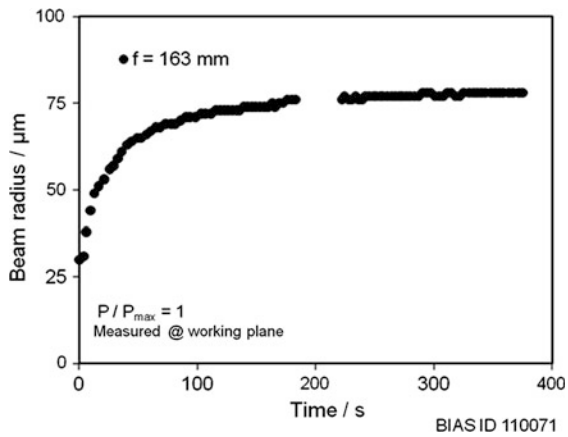


Fig. 7.18 Beam radius development over time

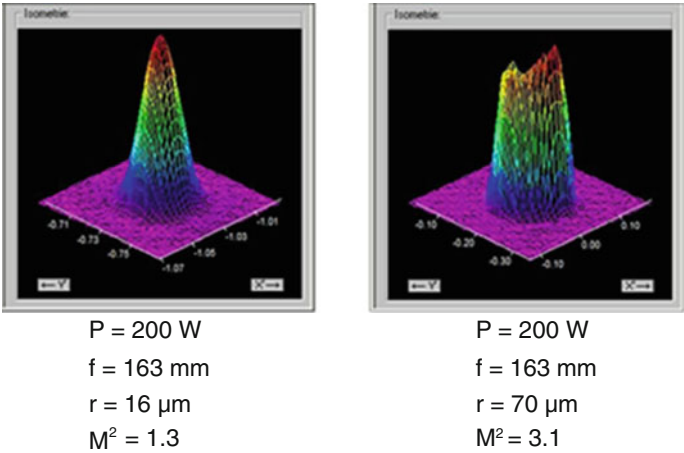


Fig. 7.19 Intensity distribution without and with thermal lensing effect

down. By this effect a thermal cycle occurs in the optics that causes the spot at the surface to change size in cycles.

The intensity of the laser beam is calculated by dividing the laser power by the irradiated area, assuming that the intensity is homogeneously distributed in the laser spot. As a consequence the change in intensity is proportional to the inverse of the squared spot diameter. How strong the effect of thermal blooming can act on the intensity distribution of the laser beam can be seen in Fig. 7.19. Moreover, the intensity profile different from the original Gaussian profile.

The effect of thermal lensing can be reduced by using high quality quartz glass optics. Figure 7.20 shows a typical optical setup of an SLM machine. The laser beam is guided with an optical fiber and collimated with a collimator lens.

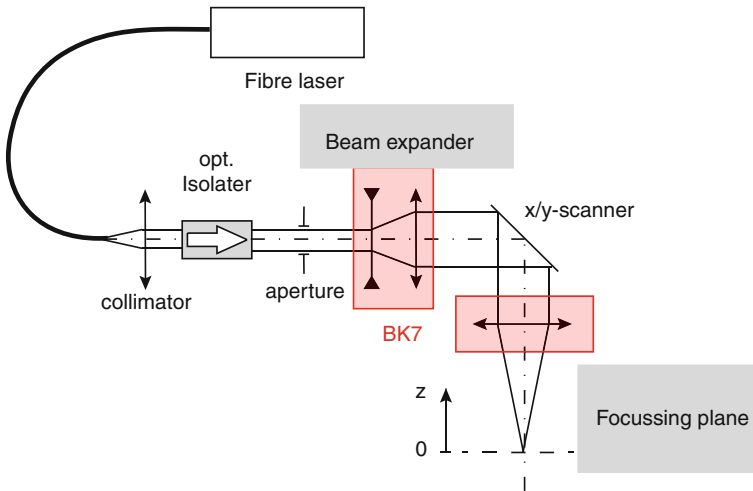


Fig. 7.20 Typically optical setup of an SLM device

Afterwards the laser beam goes through an optical isolator. The isolator blocks possible back reflections. These back reflections can damage the end of the optical fibre. After the optical isolator, the beam is passed through a beam expander consisting of a divergent and a convergent lens. One lens is connected to an electrical motor. A shift in this lens is used in order to adjust the laser spot in the z-direction in the process chamber. Two mirrors guide the expanded beam into an f-theta lens system that focuses the beam. A replacement of the beam expander and the f-Theta lens system by quartz glass optics decreased the thermal lensing and thermal blooming effect significantly. It was possible to reduce the spot size down to 50 μm continuously. With a layer thickness of 50 μm the spatial resolution is similar in the lateral plane and also vertically.

Another application of the rapid tooling technology is the possible generation of tools with adapted integrated cooling channels. In forming processes the friction between the part and tool can cause a significant increase in temperature. The friction strongly depends on the geometrical conditions of the forming process. This causes heterogeneous temperature distributions in the tool. Cooling channels in tools can reduce the temperature on the surface. SLM is an interesting technology for the generation of tools with a fully 3D-adapted cooling channel. These cooling channels can be placed in 3D with nearly complete freedom. After the consolidation of the layer, the powder remained in the cooling channels can be blown out. Another application for the cooling channel is the usage of cooling channels in injection molding tools. The cycle times of part production can be reduced by this strategy.

7.2.2 Design Rules for Micro Parts

In order to build a micro tool the following steps have to be followed. In Fig. 7.21 a possible process chain is visualized.

After the generation of the CAD model the 3D-part should be checked regarding the following points:

- overhanging structures
- minimal wall thicknesses
- bonding to the substrate
- grinding allowance.

At the overhang it is possible that the heat conduction into the powder bed below is weak and hence the melt pool size might increase. This can result in a higher maximum temperature. As a consequence the cooling time will be modified which results in a different microstructure. From a change in the microstructure, locally different mechanical properties can be the result. In addition, the direct consolidation in a loose powder bed always has a negative effect of the surface quality towards the bottom. In order to avoid these problems one can change the general design, include support structures and/or use a temperature controlled laser system.

The minimal wall thickness has in general the same disadvantages as overhang structures. Heat conduction is reduced towards the side. Moreover, the problem might be thermal distortion since when the wall is very thin, the overall structural stiffness decreases. It is suggested avoiding thin wall structures altogether. The minimum thickness should be in the order of 200 μm .

Two types of bonding towards the substrate exist. The first is the direct solidification on the base plate. The second possibility is the usage of support structures. The usage of support structures allows the easy separation of the tool from the base plate after generation. If strong residual stresses occur during fabrication the tool might distort and lift from the base plate. In the case of a direct

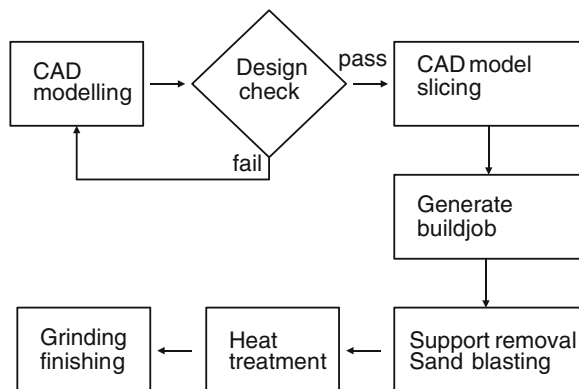


Fig. 7.21 Possible process chain for the fabrication of a micro forming tool by SLM

bond cracking can occur. If the tool lifts off during fabrication, the build job should be aborted. Otherwise damage might occur on the wiper during powder placement.

A grinding allowance should generally be included in every building job. Hence, surface roughness and thermal distortion can be compensated. Moreover the volume changes slightly due to heat treatment. The surface roughness is also influenced by the design parameters. In order to achieve a low surface roughness the powder should consist of small particles, the layer should thin and the functional surfaces should be horizontal. The thermal distortion is a consequence of the temperature gradient. The temperature gradient in the z-axis can be reduced by heating the base plate. Moreover heating has the advantage to slightly decrease the cooling speed. The volume increase by the heat treatment follows from the martensitic phase transformation.

Before the design is completed the following specific aspects of the SLM process have to be taken into account. The powder has a typical size distribution of $50\text{ }\mu\text{m} \pm 10\text{ }\mu\text{m}$ which limits the layer thickness. Typical layer thicknesses between $30\text{ }\mu\text{m}$ and $50\text{ }\mu\text{m}$ can be realized. This limits the axial resolution and causes the thickness of the slices. The lateral resolution is limited by the melt pool size. This follows the laser spot size. The build-up rate is presently between $2\text{ and }10\text{ mm}^3/\text{s}$.

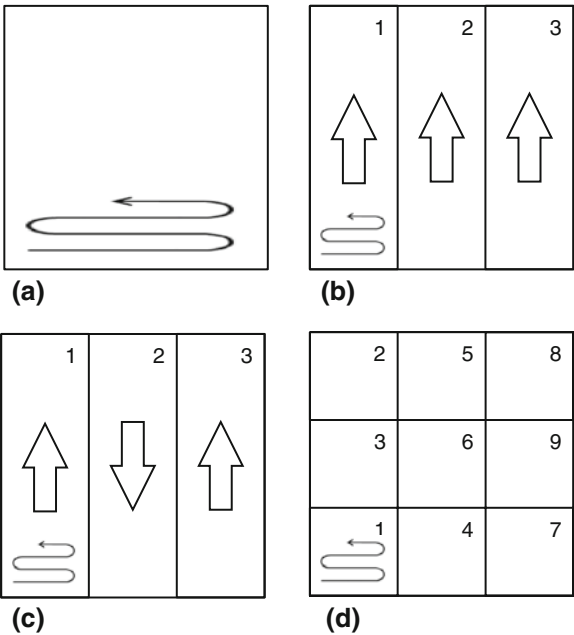
If the CAD model has passed the design check, the geometry has to be sliced. During slicing the orientation in the buildjob and the layer thickness should be specified. After slicing the process parameters and the trajectories have to be defined. A high density of the components should always be aimed for. For this reason, a comparatively high energy density should be used. This is achieved by a high laser power. A high laser power results in strong temperature gradient and hence comparably high thermal distortion. For this reason the scanning strategy is an important parameter possibility to reduce the accumulated residual stress. The following scanning strategies are common (Fig. 7.22).

The simplest strategy is the meander strategy. In this strategy the laser beam scans periodically from one end to the other. The lines are consolidated side by side. The next complex step is to divide the meander tracks into stripes. These stripes can be arranged unidirectionally or bi-directionally. A further segmentation leads to the checker board strategy. In the checker board strategy the area is divided into sub areas. These sub areas are scanned in a meander manner. The order in which the areas are treated can be chosen arbitrarily. However, it was evaluated in [112] that stresses accumulate in the different planes. This accumulation is least if the checker board strategy is applied because accumulated stress is not built in the loose powder bed in-between. It has to be taken into account that a wall might be built up at every turn around of the laser track. This leads to a larger waviness of the surface in case of the checker board strategy.

After the removal of the loose powder bed in the SLM machine the tool should be cleaned of overspray powder. This should be done by sand blasting. Moreover, the support structures, e.g. below overhang geometries, should be removed.

In order to get the mechanical properties desired, like high hardness and high wear resistance, a heat treatment is often mandatory. A hardness of 63 HRC was achieved with an SLM generated part and an additional heat treatment. Besides the

Fig. 7.22 Scanning strategies: **a** meander; **b** meander stripe (unidirectional); **c** meander stripe (bi-directional); **d** checker board



hardness, the advantage of the process is the microstructure. The carbides and grains are comparably small. Moreover, the microstructure is free of segregation. The tool steel has the following chemical composition (Table 7.4).

The heat treatment consists of three steps. Directly after generation the parts are stress relief annealed in an oven. In order to avoid oxidation the oven is flooded with inert gas. The hardening treatment is performed in a vacuum oven. The quenching is done by flooding the chamber with nitrogen. After hardening the part is tempered 3 times (Table 7.5).

After heat treatment the part has to be finished by milling technology. Special challenges and methods which have to be applied when micro forming tools are addressed in Sect. 7.4.

Table 7.4 Chemical composition of X110CrMoV8-2

X110CrMoV8-2	wt%
C	1.04
Cr	7.95
Mo	2.12
V	0.41
Si	1.08
Mn	0.22
Al	1.04
Nb	0.04
N	0.013

Table 7.5 Heat treatment for SLM generated X110CrMoV8-2 tools

Stress relief annealing	Hardening	Tempering 3 times
Inert gas	Vacuum	Inert gas
500 °C	1080 °C	550 °C
3 h	10 min	2 h
	Quenching	
	N ₂ @ 6 bar	

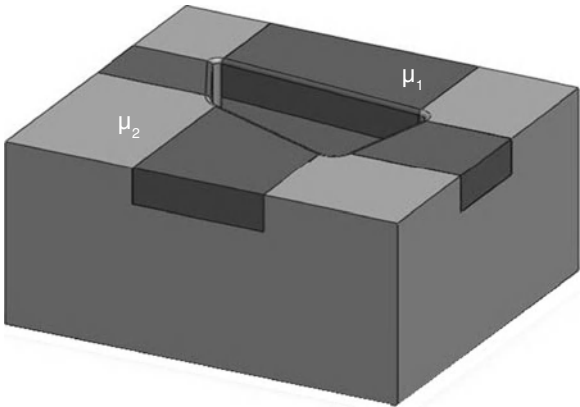
7.2.3 Multi Material Processes

An interesting extension of the SLM process for adapted micro tools is the multi material process. In this case two or more types of powder are used in order to build parts. The charming aspect is the possibility to generate either concentration gradients or rapid changes in materials. By placing different materials at different positions it becomes possible in rapid tooling technology to generate tools with tailored properties. Hence, it becomes possible to locally adapt the properties.

One approach is the generation of a tailored tool for the production of deep drawing. In general the material flow into the deep drawing die depends on the chosen geometry. In macro scaled deep drawing technology the flux can be influenced locally by applying spatial dependent lubrication or constructing a local clamping of the sheet. These possibilities are not transferable to micro scaled forming processes. The approach is to generate a micro forming die with areas of different materials to achieve different localized friction coefficient between the sheet and the tool (Fig. 7.23).

For the generation of this tool the SLM process was chosen. However, even if the resolution is sufficient, a few challenges are still valid. The diameter of the punch anticipated in this project is in the order of 1 mm. Hence, the blank for the forming process is about 2 mm in diameter. The functional surfaces of the tools that are in contact with the blank have the same size as the blank. In order to only build the important structures, the part could also be built on a platform, while

Fig. 7.23 Micro deep drawing die with locally adapted properties



using the platform as part of the object itself. This method reduces the amount of powder material and reduces the laser manufacturing time. This type of process is called hybrid SLM.

Moreover, when welding dissimilar materials it is well known that stresses occur caused by differences in the thermal physical properties of the two materials. E.g. a difference in the thermal expansion coefficient could cause compressive stresses on one side and tensile stresses on the other. A specially adapted time temperature regime can sometimes solve this problem. It is possible to preheat the powder bed by a heater in the building platform. Moreover, a scan strategy could be adapted individually in order to reduce the tensile stress.

In addition, if a hybrid structure is produced, the modification might only be necessary in the final layers in order to modify the surface properties. The amount of layers containing another material as the surrounding material should be as low as possible. Thus, it can be possible to reduce the accumulated stress caused by the different properties like the thermal expansion coefficient. As shown before, the problem of asymptotically layer height becomes important if only a few layers are deposited. The equation of the asymptotic layer height (Eq. 7.1) should be taken into account for the calculation of the correct energy deposition. Moreover, the powders should have about the same particle size distributions in order to have similar densification ratios.

In order to achieve a defect-free connection between the two materials, both have to be molten in the process. In this case a weld like bonding can be realized without pores or other defects. The complete dilution of the two materials on the interface is necessary, but complete dilution one layer with the layer below should be avoided. The dilution should be kept as low as possible but as much as necessary. It results that the resolution of the possible placement of the two different properties is a bit less compared to the lateral resolution. In general it can be assumed that the material with the lower melting temperature has to be melted after the material with the higher melting temperature. Otherwise the dilution is difficult to be kept low.

Moreover, in some materials intermetallic phases can result if the solubility is limited. These intermetallic phases are often very brittle. These can lead to the formation of cracks when the tool is under load. Moreover, micro forming technology is a technology used for a large amount of parts. For this reason it is necessary for the material to be fatigue resistant for high cycles. Intermetallic phases can have a negative effect on the endurance limit.

Metal matrix composites (MMC), e.g. WC-Co, can be used for the production of tools. Commonly the process for generating WC-Co parts is a liquid phase sintering process. With SLM, only the Co binder is melted and WC particles are distributed in the binder. The hard particles in MMC have the advantage of increasing the overall hardness of the part. In order to have a large homogeneity for the hardness, the hard particles should be smaller in comparison to the geometrical structures. In micro forming a typical radius of the die is 120 μm . The WC particles should be much smaller than the radius and thus smaller than the average particle size, which is about 45 μm . This can be achieved by using

agglomerated WC-Co powder particles. The particles are generally of about 45 μm in size. Inside, the particle consists of WC edgy-shaped particles with an average size of 1 μm . Within the Co matrix a typical mixture of 88:12 WC:Co is present. By processing this type of material it often results that some of the agglomerated sintered particles only partially melt, which causes a heterogeneous distribution of the properties. This might be avoided by multiple melting of the layers, in contrast to low dilution.

Moreover, it was shown in [99] that a NiBSi matrix can be used to embed Cr_3C_2 or WC-Co. It is clearly shown in the metallographic cross-sections that the agglomerated sintered WC-Co particles partially remain in the matrix as unmolten particles. In addition, it was shown for the case of Cr_3C_2 that the ratio between primary and secondary carbides is dependent on the process parameters.

Another possible approach for the generation of an MMC is the in situ phase formation. The basic concept is that a powder mixture of two materials are pre-mixed and used in SLM. After melting, a new phase is formed consisting of an intermediate phase of two or more elements from both powders. E.g. it was shown in [34] that a TiC reinforced Ti-Al matrix can be synthesized by using Ti, Al and graphite powder. The first step was the mechanical alloying in a 6 planetary high energy ball mill in an argon atmosphere. It turned out that nano crystalline TiC has already been generated by the process of mechanical alloying. The second step was the selective laser melting of this powder. The microstructure was still fine afterwards but small grain growth occurred.

7.3 Laser Based Subtractive Methods

7.3.1 Laser Milling

Helgi Diehl

7.3.1.1 Introduction and Definition

The sustainable industrial production of micro components of intricate shape, high contour accuracy, surface quality and preset material characteristics is strongly governed by the availability of appropriate, reliable, systematically designed and manufactured tools. Tool making for micro metal forming deals with issues such as tool shape, surface and bulk properties on submillimeter scale and is, therefore, highly challenging. In this sense, very few approaches can be considered. Application of a laser beam is one of them and belongs to non-conventional methods of tool making. Up to now, laser material processing is not entirely understood. Moreover, it is presumed to be expensive owing to the high investment and maintenance costs. However, the wide implementation of lasers in modern production lines as well as a growing interest of users in novel laser systems

demonstrates the high esteem of the unique laser properties. In machining applications, the focusability of the laser beam is one of them. This feature facilitates high irradiances and hence allows high local energy deposition; unlike conventional thermal sources like the welding torch or mercury arc lamp. This allows, in particular, quick, accurate and controlled material removal for tool machining, i.e. laser micromilling.

Laser milling is a subtractive manufacturing process which utilizes the laser beam as a tool for the controlled removal of material from a substrate. It allows the fabrication of parts from a wide range of tool materials, including metals and insulators. Similar to conventional milling, the focused laser beam is moved relative to the workpiece, usually via a scanning system, and creates three-dimensional shapes in the material. The chip generation known from mechanical machining is not necessary due to the nature of light-matter interaction and possible mechanisms of material subtraction.

7.3.1.2 Laser Subtracted Chip

Subtraction of the material by laser radiation proceeds via mechanisms such as vaporization, melt expulsion, coulomb explosion of ions, homogenous boiling, spinodal decomposition, spallation, etc. Even though classical chip generation is usually not present, relations known from mechanical milling may be, however, useful. In conventional machining, high accuracy correlates with the diminution of the chip subtracted. Similarly, the “chip” removed by laser radiation must be minimized in order to facilitate high contour accuracy. The volume of the laser subtracted chip (see Fig. 7.24b)—characterized by lateral radii $r_{s,x}$ and $r_{s,y}$ as well as by the chip thickness h_s —is due to the volume affected by radiation which usually extends over the boundaries of the excavated volume (see Fig. 7.24a). The affected volume is a product of the laser affected area $A_{a,xy}$ times the laser affected depth h_a . While the first results from the laterally affected radius $r_{a,x(y)}$ which is a sum of the laser spot radius r_w and the thermal penetration radius r_{th}

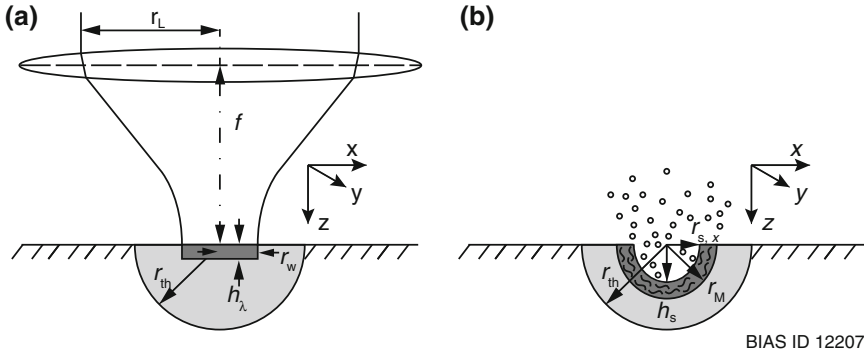
$$r_{a,x(y)} = r_w + r_{th}, \quad (7.2)$$

the latter is determined by the optical penetration depth h_λ and the radius r_{th}

$$h_a = h_\lambda + r_{th}. \quad (7.3)$$

In the case of a focused laser operated in a fundamental single mode regime, r_w denotes the laser beam waist radius as a radius at which laser center intensity decays by a factor of $1/e^2$ which is given by

$$r_w = \frac{\zeta \lambda f}{\pi r_L}, \quad (7.4)$$



BIAS ID 122071

Fig. 7.24 **a** Definition of the material volume affected by a focused laser light (*dark gray*) and the subsequently expanded HAZ (*light gray*). f , r_L , r_w , h_λ and r_{th} denote the lens focal width, the beam radius at the lens, beam waist radius, optical and thermal penetration lengths, respectively. In metals, $h_\lambda \ll r_{th}$ is usually fulfilled. **b** Definition of the subtracted material “chip” (*circles*), melted layer (*dark gray*) and HAZ (*light gray*). Hereby, r_M , $r_{s,x}$ and h_s and denote the radius of the recast layer, the lateral radius and the depth of the subtracted chip, respectively

whereby λ is the laser wavelength, ζ the beam quality factor, r_L the beam radius at the lens and f the lens focal width. Small features can be illuminated by applying laser systems with short wavelengths and low ζ factors, like frequency converted solid state lasers as well as optical setups providing large convergence angles and, hence, high r_L/f ratios. Unlike the lateral extension of the incident beam, the optical penetration length or the *skin layer* h_λ is determined as

$$h_\lambda = 1/c_\alpha = \lambda/4\pi k_d, \quad (7.5)$$

with c_α denoting the spectral absorbance (exponential attenuation) and k_d the damping constant in the material. The skin layer h_λ ranges for most steels and cemented metal carbides owing to high damping k_d in the order of tens of nanometers. The main contribution to the laser affected depth in Eq. (7.3) mostly originates, therefore, from the thermal penetration radius r_{th}

$$r_{th} = 2\sqrt{\kappa t_D}, \quad (7.6)$$

which is a square root function of the material-dependent thermal diffusivity κ and the laser pulse length t_D . More generally, the time t_D should be defined as a time of light-matter interaction or a dwell time. This is reasonable in milling applications with a significant relative motion of a sample and a laser beam. Hereby, the effective pulse acting on a chip volume is shortened. In processing applications with a continuous wave (cw) laser moved over the sample surface with a constant feed rate v_f , one may assume the effective dwell time as $t_D = 2r_w/v_f$. Equation (7.6) obviously concludes that the reduction in pulse time t_D results in a decrease of the thermally affected volume, i.e. heat affected zone (HAZ), which causes structural modifications in the material and is usually attempted to be avoided. Furthermore, short t_D consequently allow the subtracted chip to be

downsized (see Fig. 7.24b). Although true in a long pulse range, the straightforward scaling down of t_D below a definite value does not lead to any further reduction in the HAZ. This limit is defined by the material specific electron–phonon scattering time t_{e-ph} which is in the range of $10^{-12} - 10^{-10}$ s for metals and may reach 10^{-6} s in insulators. For $t_D < t_{e-ph}$ the HAZ and the molten layer stabilize at some material-dependent value. Another determining factor in Eq. (7.6) is the thermal diffusivity κ (Table 7.6)

$$\kappa = \frac{K}{\rho c_p}, \quad (7.7)$$

whereby K is the thermal conductivity, ρ the density and c_p the specific heat capacity. This value depends strongly on the current material properties and is high in solid conducting materials. This ensures, in particular, that r_{th} is the determining contribution in the definition of the laser affected depth h_a .

In contrast to metallic alloys and carbides, in dielectric tool materials like quartz, fused silica, corundum/sapphire and alumina ceramics the thermal diffusivity κ is low. As a consequence, the affected depth is mostly governed by the skin layer h_λ (dark gray in Fig. 7.24a) which is strongly dependent on the wavelength. For instance, fused silica absorbs UV light (low h_λ) owing to electronic band absorption and IR radiation due to vibrational band absorption being, however, transparent in the intermediate VIS range (large h_λ). Moreover, h_λ is a function of the laser intensity and dwell time. Indeed, non-linear absorption of the intense radiation as well as the absorption enhancement owing to in-process emerging incubation centers may shorten the effective optical depth. Among dielectrics, diamond builds a special case since it possesses a large κ owing to its non-electronic character of thermal conductivity. Therefore, in diamonds both optical and thermal penetration depths should be accounted for when defining the laser subtracted chip size.

7.3.1.3 Method of Choice

Precise tool making requires controlled chip generation on micro scale providing high contour accuracy and low surface roughness. Hereby, minimal structural modifications in the bulk material should be pursued. For this purpose, reasonable thermal impact originating from the HAZ should be avoided. Besides precision, in industrial manufacturing cost-effectiveness plays an important role. Therefore, removal approaches and processes revealing high volume material removal rates (MRR) are desired. Unfortunately, a universal method for controlled chip subtraction with both high precision and high processing efficiency is not existent. Therefore the user has to balance the quality requirements in every singular milling application against the anticipated efficiency and make a decision. The following relationships are considered, however, as generally applicable and may serve the user as a roadmap.

Table 7.6 Minimal feature diameters $d_{\min}$, average surface roughnesses R_a and volume removal rates $\dot{V}$ reported in bulk tool materials by applying various laser milling approaches

Material	Laser milling approach	Laser, t_D , harmonics	$d_{\min}$ [μm]	R_a [μm]	$\dot{V}$ [mm^3/s]	References
Stainless steel	Non-reactive gas jet	Nd:YAG, cw	200	< 2	34	[33]
Steel 25CrMo425	Non-reactive gas jet	CO ₂ , pulsed	200	< 2	8	[79]
Stainless steel	Grit blasting	CO ₂ , cw	1000	15 - 42	70	[58]
Stainless steel	Reactive gas jet	Nd:YAG, cw	500	4	< 78	[33]
Stainless steel	Reactive gas jet	Nd:YAG, cw	270		36	[43]
Cold working steel	Reactive ambient	Nd:YAG, cw	100	10*	3×10^{-3}	[41]
Alumina	Controlled fracturing	Nd:YAG/CO ₂ , cw	500	60	0.15	[102]
Stainless steel	Long pulse ablation	Nd:YAG, 10 μs	40	2.4		[75]
Stainless steel	Long pulse ablation	Yb: Fiber, 20 μs	40	10.7	2.5	[35]
WC-10Co	Long pulse ablation	Nd:YAG, 100 ns	25			[26]
Fused silica	Short pulse ablation	Excimer, 33 ns	380		3×10^{-3}	[42]
Steel	Short pulse ablation	Co vapor, 10 ns		0.3**	6.5×10^{-3}	[50]
CVD-diamond	Short pulse ablation	Co vapor, 10 ns			1.5×10^{-3}	[50]
WC	Short pulse ablation	Nd:YAG, 63 ns			5×10^{-2}	[37]
Steel	Ultrafast ablation	Nd:YVO ₄ , <15 ps, 3rd		0.3		[50]
Alumina	Ultrafast ablation	Nd:YVO ₄ , <15 ps, 3rd		0.25		[48]
Fused silica	Ultrafast ablation	Nd:YVO ₄ , 8 ps, 3rd		0.3		[49]
WC-Co10	Ultrafast ablation	Ti:Sapphire, 150 fs			2×10^{-4}	[26]
Synthetic diamond	Ultrafast ablation	Ti:Sapphire, 150 fs			2×10^{-5}	[26]
Sapphire	Ultrafast ablation	Ti:Sapphire, 164 fs	4		$0.3 - 1.5 \times 10^{-5}$	[78]
Stainless steel	Liquid chemical jet	Yb:Fiber, cw	10	0.3	1.6×10^{-4}	[96]
Stellite 21	Liquid chemical jet	Yb:Fiber, cw	10	0.3	1.6×10^{-4}	[93]
Titanium	Liquid chemical jet	Nd:YAG, cw	10	0.1	1.6×10^{-4}	[94]

* The data are presented for Rz

** Sacrificial layer has been applied

From a material point of view, structures with a large value of absorbance c_α and small value of diffusivity κ are beneficial as they aid fine chip generation and, hence, the achievement of small features inhibiting the energy dissipation in the surrounding area and preventing structural modifications. However, in tool making one has to handle every certain basic raw material with given characteristics. Therefore, parameters of the laser system as well as those of assisting media become of crucial importance.

When considering ablative methods (see Sect. 7.3.2), the most important parameters are the laser wavelength λ , laser intensity I , the pulse length or dwell time t_D and last but not least the pulse frequency or laser repetition rate f_{la} . Often in literature a composite parameter called fluence $\phi \propto It_D$ is considered. Generally, short pulse lengths allow the intensity to be increased by bunching the output energy and the dwell time t_D to be diminished. Ultrashort pulses ($t_D < 100$ ps) therefore facilitate the quick deposition of highly intense light energy into the surface matter with low structural modifications, owing to the minimal possible HAZ. For instance, gentle milling of metals was shown to be advantageous with picosecond lasers. For dielectrics, femtosecond laser pulses appear to be a kind of a prerequisite to raise the intensity towards the levels necessary for a subtractive processing which hence occurs over nonlinear phenomena. Variation in the wavelength is a further instrument for more accurate as well as more efficient tool processing. The application of lasers with shorter wavelengths to conducting materials are in particular more plug efficient due to the lower reflectance of those in the VIS and UV spectral range as compared to IR illumination. In dielectrics, the UV laser light can even cause photolytic ablation of the material via the direct bond breaking without significant thermal damage, often referred to as “cold” ablation. Last but not least, light of shorter wavelengths is better focusable due to Eq. (7.4), leading to smaller lateral dimensions of the optically affected region and boosting both the intensity and temperature in the focus. The intensity in the focus is recommended to be chosen slightly above the material specific threshold. In this case, the ablated chip remains small; also the surface roughness and the structural modifications are lower compared to intense illumination due to more confined melted area and the HAZ (compare $I^{(2)}$ and $I^{(1)}$ in Fig. 7.30, respectively). Tuning low intensities, short wavelengths and short pulse durations usually results in a rise in milling precision; however, it decelerates the economic feasibility due to generally low MRR. Application of high laser repetition rates f_{la} is a tool to compensate for low throughput. These are to be chosen as high as possible, however are typically up to 20 MHz. Burst modes above this value are too fast to allow the ablated matter to leave the ablated area and, hence, may lead to losses in milling quality.

Even though modern laser systems allow high definition material subtraction with high removal rates by means of ablation, the highest MRR can still be achieved by applying hybrid laser milling systems as discussed next. Hybrid methods are strongly influenced by additional factors such as the constitution and aggregate state of the involved medium, its pressure and—in cases of a directed jet—the angle of incidence. An overview of the different approaches for the

milling of various tool materials is presented in Table 7.1, revealing three important characteristics: minimal lateral feature diameter d_{min} , average surface roughness Ra and volume MRR $\dot{V}$. Their concise description, including the most relevant processes, their principles, physical background, processing methods and application data is the subject of the following review.

7.3.1.4 Laser Milling Via Non-Reactive Gas Jet

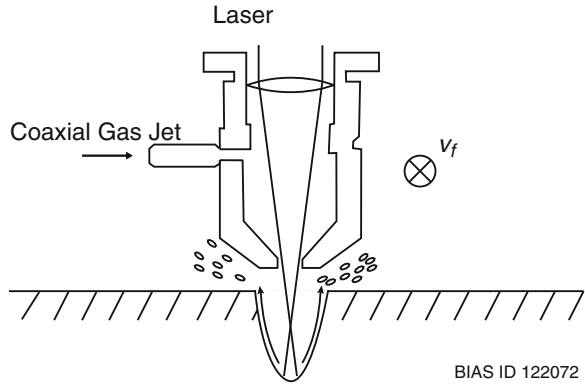
Simultaneous application of laser radiation and a high pressure gas jet is a hybrid approach in laser milling, often denoted abrasive laser machining. It is characterized by high MRR and provides, for certain demands, a sufficient quality. While the cw lasers are considered as heat suppliers for selective melting, the jet pressure is concurrently used to expel the liquid-phase and, hence, to locally subtract the material (Fig. 7.25). The advantage of the approach is its plug efficiency, since much less energy per volume is required than in laser induced vaporization. Indeed, the energy needed to vaporize 1 gram of iron ranges at around 6.3 kJ, the energy required to melt it—about only 0.3 kJ!¹

When laser radiation of peak intensity I_0 strikes a tool material surface, a part of it proportional to $c_R I_0$ is lost by reflection, another part is absorbed in the material or/and is transmitted. For most metals, the absorption coefficient c_α is in the range $10^7 - 10^8 \text{ m}^{-1}$ which means that the absorption takes place at the surface within a skin layer of $10 - 100 \text{ nm}$; for bulk samples no transmission occurs. Therefore, the absorbed intensity I_{ab} at the center of the irradiated area is described by $I_{ab} = (1 - c_R)I_0$. Due to high reflection of infrared irradiation by solid metals, an effective coupling into the material is usually mediated via plasma formation ($I_0 > 2.5 \times 10^7 \text{ W/cm}^2$) or at least after a characteristic time necessary to heat/melt the surface layer and, hence, reduce the reflectivity. For instance, the reflectivity of steel is reduced by a half after 600 ns when irradiating with CO_2 laser radiation at $t_D = 200 \text{ ns}$ and intensity $I_0 = 1.5 \times 10^8 \text{ W/cm}^2$ [80]. Therefore, the application of long pulses or scanning cw radiation appears to be advantageous in the sense of effective energy coupling into the medium. The goal of the approach is the controlled melting of a chip with defined dimensions. From the energy balance, the maximum depth $r_{M,max}$ of the chip melted may be estimated from

$$r_{M,max} \approx \frac{I_{ab} t_D}{\rho (\Delta H_M + c_p [T_M - T_0])}, \quad (7.8)$$

¹ The data only represent the enthalpies of vaporization and melting. The total energies including the heating of the material and relevant enthalpy contributions are, hence, 7.8 and 1 kJ, respectively.

Fig. 7.25 Basic principle of tool milling by laser-gas jet approach. v_f denotes the feed velocity vector



whereby T_M and ΔH_M denote the melting temperature and melting enthalpy, respectively, and T_0 is the initial temperature or/and the temperature of the ambient medium. Tuning the absorbed fluence $\phi_{ab} = I_{ab}t_D$ as a composite may deliver both insufficient processing conditions and milling results. Indeed, long t_D and intensities that are too low lead to extended HAZ in the material and, hence, usually unwanted structural modifications. Otherwise, too short dwell times but high peak intensities I_0 lead to overheating and vaporization of the surface, which are not plug efficient processes. The optimum laser parameters t_D and I_0 can be found from the condition $r_{M,max} \approx r_{th}$ which are then given by

$$t_D \approx \frac{r_M^2}{4\kappa} \quad (7.9)$$

and

$$I_0 \approx \frac{2\rho(\Delta H_M + c_p[T_M - T_0])}{1 - c_R} \sqrt{\frac{\kappa}{t_D}}, \quad (7.10)$$

respectively. As an example, the optimum dwell time and intensity for melting a chip in iron with a depth of $10 \mu\text{m}$ can be estimated as $t_D \approx 1 \mu\text{s}$ and $I_0 \approx 2 \times 10^7 \text{ W/cm}^2$. Unfortunately, industrial lasers with such output parameters are not present today. Alternatively, high power cw lasers with high feed rate v_f or pulsed solid state solid state lasers pumped by appropriate power supply units provide intensities in the range of 10^4 W/cm^2 at pulse lengths $t_D \approx 1 \text{ ms}$ suitable for the melting of a $10 \mu\text{m}$ thick layer, however at the cost of an extended HAZ. If the low extension of the HAZ is of basic importance for the user, then higher intensities at shorter dwell times are required which lead to higher energy deposition in the surface layer and its subsequent overheating. In this case, the maximum achievable depth of melting may be reduced as it correlates with the maximum allowable surface temperature, which is the boiling point. Meanwhile, the lateral dimensions of the chip are in accordance with Eq. (7.2). Geometrically, the chip melted obtains a rounded cylindrical form if $r_w^2/\kappa t_D \gg 1$ (compare with

melted zone for $I^{(2)}$ in Fig. 7.30b). In the case of the strongly opposite ratio, the melted chip reveals the form of a filled hemisphere with radius r_M . Once melted and in the absence of the vapor recoil pressure or/and plasma shock waves at high intensities, the liquid phase is retarded to leave the kerf owing to surface tension. Hereby, an external source of gas pressure with Mach numbers of, typically, around 0.2 offers an effective instrument to expel the melt off the cavity leading to chip subtraction. The alignment of the nozzle is often chosen to be coaxial with the orientation of the laser processing optics as shown in Fig. 7.25. Last but not least, concurrent expulsion is of particular significance. Indeed, after the pulse ends and in non-reactive ambient conditions the melt starts to solidify from both the bottom liquid–solid as well as from the upper liquid–gas interfaces. The experimentally determined solidification velocities in metals range from 10 to 10^3 m/s. These times have to be considered in the process layout.

Physically the process can be described by predominantly photothermal phenomena leading to melting. The electrons absorbing the light energy conduct the energy in the interior of the tool material by electron–electron scattering and transmit it to the lattice ions by electron–phonon scattering, which is accompanied by a perceptible temperature rise. According to the temperature profile, the solid liquefies and the fluidity of the medium becomes locally enhanced. The liquid phase is then abraded by mechanical forces provided by directed gas pressure waves.

In industrial tool milling, generally imposed limitations like the attainment of a high surface quality and preservation of constant mechanical properties require the avoidance of overheating. Therefore, instead of pursuing the near-net shape of the surface in one shot or in a stationary mode, the application of a scanning (focusing) technique is advantageous. The latter is based on a series of discrete overlapping paths while machining low diameter craters (pulsed mode) or narrow grooves (cw mode) in each single flow. In this case, maximum depth limitations can be overcome to some extent. Small features can be achieved by the application of short wavelengths, lasers with high beam quality and in pulsed or short dwell time regimes. Widely applied systems are multi-kilowatt continuous wave CO₂ lasers, however Nd:YAG laser systems allow much more flexible handling by guiding the beam over fibers. Moreover, they are better focusable and more efficiently coupled with metals owing to shorter emitted wavelengths. Lasers organized within glass fibers, i.e. fiber lasers, are newcomers and combine the advantages of solid state lasers with flexible beam optics, presently delivering the highest intensities in cw mode. These systems can be quickly scanned relatively to the work surface, either by moving the working head, the stage or even both, and are controlled by CNC machines. For better handling of the workpiece, long focal lengths are preferred. Short dwell times realized by quick scans (high feed rates) or pulsed radiation as well as low intensities above the melting threshold provide, in general, small features, low HAZs and low roughness [33]. Usually, higher intensities at pulse lengths below 1 ms make it difficult to avoid vaporization which, hence, increases diffraction losses impacting negatively on the machining quality. An assist gas jet is used to protect the optics from contamination and to facilitate material removal.

Hereby, assist gas type, gas pressure as well as nozzle geometry and position are of additional importance. Usually gases such as nitrogen, argon and helium are used. The process modeling is complex and can be proceeded by solving a set of balance equations for energy, momentum and mass including all relevant parameters of the laser beam and assist gas flow [46].

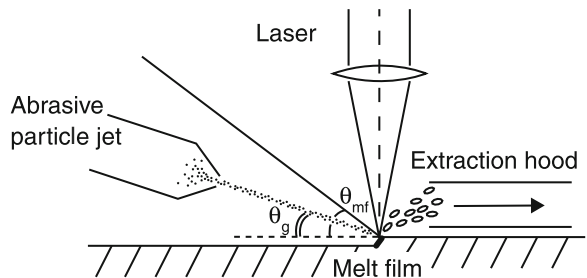
Application of a laser-jet approach to milling has been addressed by different authors (see for instance [19]). The most data on feature sizes, surface quality and removal rates originate, however, from metal cutting applications. Nevertheless, the knowledge gained may also be employed in laser milling. Machining of steels by cw Nd:YAG laser and pulsed CO₂ laser (500 Hz) with an average power of ca. 700 W have resulted in features with $d_{min} = 220 \text{ }\mu\text{m}$, $Ra < 2 \text{ }\mu\text{m}$ and $\dot{V}$ about $34 \text{ mm}^3/\text{s}$ [33] and $8 \text{ mm}^3/\text{s}$ [79], respectively. The average HAZ was found to be around $50 \text{ }\mu\text{m}$. The presented data hence favors the Nd:YAG lasers, providing higher removal rates with approximately the same accuracy. A strong disadvantage of the method is a wide extended HAZ leading to enhanced structural modifications and the low surface quality due to the recast layer and debris.

7.3.1.5 Laser Milling Via Grit Blasting

Laser melting and expulsion via grit blasting is a variation on the former approach. For melt expulsion, different abrasives based mainly on silicon carbide and alumina particles have been efficiently utilized which are delivered with a compressive gas jet to the melt region. An issue is, hereby, the angle of particle impact. In order to remove the material from the cavity, the angle of incidence θ_g should be smaller than the angle of the molten front θ_{mf} hence providing an up-lifting force (see Fig. 7.26). The extraction hood is used coincidentally to collect the raised material.

In practice, particles of $50\text{--}110 \text{ }\mu\text{m}$ size are delivered by a compressive air jet. Application of a TEM₀₁ mode of cw CO₂ laser at 550 W and a jet at 5 bar pressure to the steel 304, have yielded removal rates of $70 \text{ mm}^3/\text{s}$, $Ra = 15\text{--}42 \text{ }\mu\text{m}$ and HAZ of $20\text{--}50 \text{ }\mu\text{m}$ [58]. Variations in abrasive jet inclination has revealed that optimal melt removal can be achieved for oblique jet orientations at $\theta_g = 25^\circ$. The advantages of grit blasting are its high removal rates and relatively low HAZ,

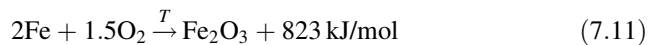
Fig. 7.26 Principle of tool milling via grit blasting. θ_{mf} and θ_g denote the angles of the molten front and grit blasting, respectively. According to [58]



however, high roughness values make a mechanical machining step necessary. Feature diameters below 1 mm are possible with this method.

7.3.1.6 Laser Milling Via Reactive Gas Jet

Reactive laser machining is another variation on the basic method, whereby a reactive gas is used instead of an inert gas. The reactive gas not only acts as an abrasive but supplies additional energy owing to an exothermic reaction. Introduction of oxygen as a jet gas leads to an enhancement of MRR originating from combined effects such as a higher absorptivity of incident light by the hot oxides, additional channel of heat input provided by exothermic reaction and lower effective viscosity (lower surface tension) beneficial for abrasion. In steels, the energy supplied by a reaction of iron with oxygen atoms is



The reaction requires a certain initial temperature which is below the boiling point. Once started and for certain processing conditions, the energy generated by the reaction may even exceed the energy delivered by the absorption of the laser beam. The concurrent ejection of the molten material from the chip volume by the gas pressure leads to material removal. Basically, the absorbed laser radiation in combination with an exothermic reaction affords enough energy to facilitate the melting and even evaporation from the surface. Both, the evaporation pressure (recoil pressure) as well as the pressure submitted by striking gas jet may lead to melt expulsion. The surface evaporation (ablation) has been shown, however, to be negligible. The affected chip volume is enlarged both optically, by the vapor scattering of the incident light, and thermally, as a result of the exothermic reaction.

Machining of stainless steel by cw Nd:YAG laser systems yielded roughness values $Ra \approx 4 \mu\text{m}$ [33], feature sizes $d_{\min} = 270 \mu\text{m}$ and removal rates $\dot{V}$ about $36 \text{ mm}^3/\text{s}$ [43] therefore revealing high processing speeds. However, great disadvantages are the modifications in material constitution due to surface oxidation, enlargement of the feature size, wide extension of the HAZ, high roughness values and striations, i.e. periodic lines appearing on the walls. Application of the method to nonmetals has revealed that the efficiency and quality of milling is generally less influenced by the type of assist gas used.

7.3.1.7 Laser Milling in a Reactive Ambient Gas

A combination of a laser and a reactive ambient gas is another hybrid approach in laser milling. Both gaseous as well as liquid phases have already been successfully introduced. When applying a combination of a low intensity cw laser and an oxygen gas, material removal occurs owing to oxide stress chip generation. An

important precondition here is that the melting point of the oxidized dross is lower than the melting point of the host material.

The initial physical mechanism is of photothermal origin. The material is exposed locally to low irradiances leading to material heating without melting. The supply of oxygen leads to an exothermic reaction of the heated material providing additional energy and altering the material constitution in the focus by oxidation. During the resolidification time of the melted oxide, mechanical forces owing to the difference in the shrinkage behavior of the dross and the host metal yield local separation of the oxidized flake from the bulk material.

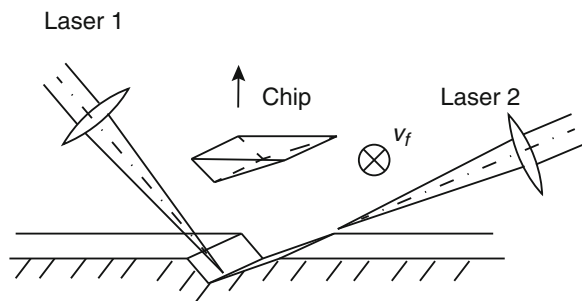
From a practical point of view, submission of oxygen can be enabled by a stage with a coaxial gas jet. The gas pressure parameter should however be optimized in order to avoid the expulsion of the melt similar to laser milling via reactive gas jet. A variation of the jet technique may be the oxygen sphere supplying the reactive molecules to the heated spot. A focused laser beam with output powers ranging around 10 W is scanned over the surface in multiple paths providing chip generation on the micro scale.

Application of this method to metals yields features with lateral dimensions in the range of 100–500 μm and depths from 20–100 μm . In particular, irradiation of cold working steel (EN X160CrMoV121) samples with the light of a cw Nd:YAG radiation at $P_{la} = 7\text{ W}$ has allowed the achievement of groove depths of 80 μm , $\dot{V} = 3 \times 10^{-3}\text{ mm}^3/\text{s}$ and $R_z = 10\text{ }\mu\text{m}$ [41].

7.3.1.8 Laser Milling by Intersecting Beams

The basic concept of milling is based on chip-by-chip subtraction. This requires either melting or vaporization of the whole volume to be removed and hence significantly raises the energy consumption. If energy efficiency is of capital importance an alternative approach with two lasers can be applied [15]. This implies a setup in which both laser beams are directed on the material surface and the incidence axis of the first beam intersects the axis of the second beam, as shown in Fig. 7.27. It visualizes how blind cuts or drills may result in the removal of chips with a triangular cross-section by their conjunction. The subtracted

Fig. 7.27 Basic principle of tool milling by intersecting laser beams. v_f denotes the feed velocity vector. According to [15]



volume obviously depends on the intersecting point, the angle of convergence and the properties of the blind kerfs or grooves produced. By elaboration of sophisticated milling strategies, material removal rates comparable to those of mechanical machining can be achieved. The surface quality and the accuracy are mostly given by the straightness and the width of the kerfs and therefore depend on the properties of the cuts and/or drills resulting from each single approach, which can be of hybrid or ablative nature.

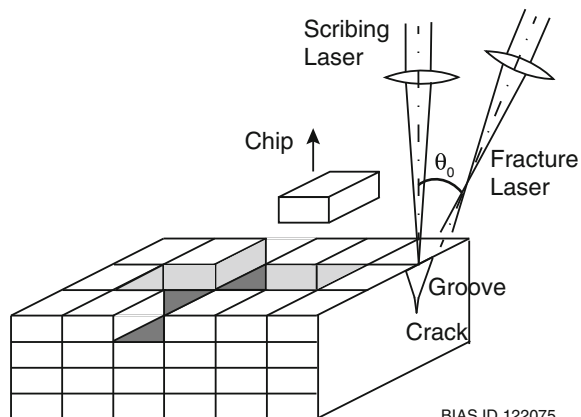
7.3.1.9 Fracture Machining

Another energy efficient method is called fracture machining and is based on controlled fracturing. First, a focused laser is used to scribe well defined grooves. In order to induce fracturing, immersion of a sample in an ultrasonic cell has been successfully applied [32]. Another possibility is the application of defocused laser radiation as a heating source [59]. The thermally generated tensile stresses have been shown to result in the formation and extension of cracks from the tip of the grooves downward. Their linkage leads to fracturing and material removal. This approach is similar to machining with intersecting beams as it is based on a breaking off of 3D chips out of the substrate and is, therefore, energy efficient. By its nature, the current method is well suited for the milling of brittle materials like ceramics and glasses where it provides high speeds and clean surfaces.

In processing it is convenient to use two lasers serially. While the scribing laser is focused on the surface structure and oriented normally, the second one, responsible for the propagation of cracks, is aligned obliquely at an angle θ_0 , as demonstrated in Fig. 7.28, and is defocused.

Applying a focused cw Nd:YAG laser beam ($P_{la} = 53 \text{ W}$) of $32 \text{ }\mu\text{m}$ beam waist diameter as a scribing tool and a $373 \text{ }\mu\text{m}$ wide cw CO_2 laser beam ($P_{la} = 44 \text{ W}$) inclined at angle $\theta_0 = 15^\circ$ for the fracture-machining of closed square elements in alumina (see Fig. 7.28), chips with dimensions below 1 mm at $\dot{V}$ of $0.15 \text{ mm}^3/\text{s}$ have

Fig. 7.28 Basic principle of tool milling by fracture machining. θ_0 denotes the inclination angle of a defocused fracturing laser. According to [102]



been extracted [102]. The achieved roughness Ra ranged at around $60\text{ }\mu\text{m}$. Improvement of milling strategies may enhance the removal efficiency. Indeed, fracture-machining applied to the edge of alumina has yielded $\dot{V} = 10.8\text{ mm}^3/\text{s}$ [103].

7.3.2 Laser Ablation

Helgi Diehl

7.3.2.1 Definition

The term laser ablation is used if material subtraction can be performed in a vacuum or in an inert ambient medium and requires the direct absorption of laser light by the material. Ablation occurs, in general, at laser intensities above the material vaporization threshold I_V ranging above 10 kW/cm^2 . In order to submit such irradiances a pulsed mode regime is usually required. In the following, three pulse regimes—long pulse (100 ns–100 ms), short pulse (100 ps–100 ns) and ultrashort pulse (100 fs–100 ps) ranges—are distinguished. Within each approach specific laser setups are tailored to meet the challenges of controlled and precise laser milling. Indeed, tuning effective dwell time but also variation in radiation intensity and wavelength have a strong impact on the root and hence the result of the milling process.

7.3.2.2 Laser Ablation with Long Pulses

Laser radiation with a pulse length from 100 ms down to 100 ns and intensities above 10 kW/cm^2 may result in surface evaporation. In metals, the vaporization usually proceeds after passing a melting phase. Then, the evaporated material imposes a recoil pressure on the surrounding which, acting on the melting pool, leads to the hydrodynamic flushing of the melt from the cavity providing an additional channel of removal.

Surface evaporation is the basic process of material ablation. The advantages are the reduction of recast layer, lower HAZ and better surface quality. In metals and conducting ceramics, this process calls for shorter wavelengths in order to couple the radiation efficiently into the matter and to produce smaller lateral features. Commonly used systems are the Q-switched (pulsed) Nd:YAG lasers with wavelength of $1.064\text{ }\mu\text{m}$, however CO_2 TEA lasers are also applied. For visible and near infrared region, the intensities at which the evaporation is a dominant process lie within the range from 1 to 150 MW/cm^2 . While in the lower intensity regime the melting becomes prevailing, at intensities above the upper threshold the strong absorption of incident light by the vaporized material leads to

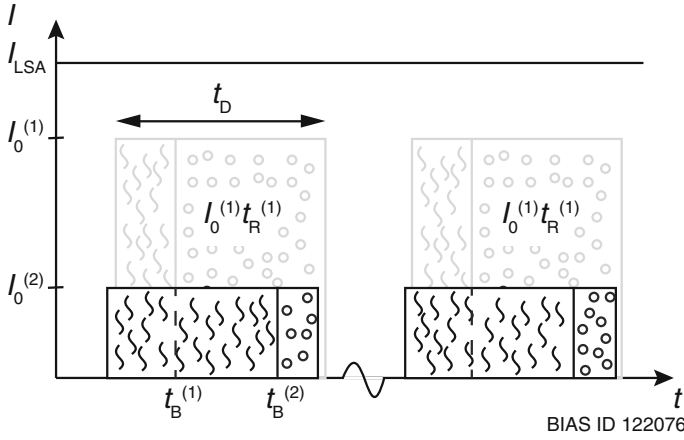


Fig. 7.29 Allocation of laser energy (intensity I) for melting (wave) and evaporating (circles) of material in time. I_{LSA} denotes the threshold intensity of plasma generation

generation of plasma. The latter interacts strongly with the high intensity light blocking the incident radiation, diminishing the surface impinge and hence the ablation rate. Additionally, plasma shock waves can result in material modification zones owing to shock hardening. Therefore, in long pulse ablation the upper limit should not be crossed.

Volume MRR by evaporation can be quantitatively estimated from the energy balance approach. The latter is based on a simple approximation implying two stages: the melting and the removal by vaporization (see Fig. 7.29). Theoretical calculations based on specific model assumptions allow calculation of the boiling time t_B , i.e., the time required to melt the surface and reach the vaporization stage

$$t_B = \frac{\pi K \rho c_p}{4 I_{ab}^2} (T_B - T_0)^2 \quad (7.12)$$

whereby T_B denotes the boiling temperature. As an example, the time needed to reach the boiling point in iron by illuminating it with common lasers at intensities $I_0 = 1 \text{ MW/cm}^2$ and $I_0 = 10 \text{ MW/cm}^2$ is estimated to be $18.6 \mu\text{s}$ and 186 ns , respectively [80]. Within the energy balance model, the uniaxial removal (drilling) velocity v_z is given by

$$v_z = \frac{I_{ab}}{\rho [c_p (T_b - T_0) + \Delta H_M + \Delta H_V]} \approx \frac{I_{ab}}{\rho \Delta H_V}, \quad (7.13)$$

whereby ΔH_V is the material-dependent enthalpy of vaporization, accounting by approximation for the sum in the denominator brackets and determining the energy efficiency of the ablation. In the case of “atom-by-atom” vaporization, the enthalpy ΔH_V boosts the energy required to vaporize 1 gram of atoms, allowing, however, the removal of smaller chips. For ablation in the form of droplets or large

fragments, the average removal energy per atom decreases, raising the energy efficiency of the process. The applicability of the introduced model has been tested by conducting a number of experiments. Those performed on steel 316 and tungsten carbide have verified the theoretical estimations, demonstrating that time t_B is reached 27 times quicker in steel than in WC samples [36]. When applied to tungsten carbide, it is noted that the radiation of Nd:YAG laser with $I_0 < 10 \text{ MW/cm}^2$ and t_D as long as 160 ns is not even capable of starting ablation. This result is, however, plausible within the theoretical model: t_B exceeds t_D leaving no time for the removal stage. Under appropriate fixed conditions, within removal time $t_R = t_D - t_B$ the steels are ablated 1.3 times faster than the WC samples. With respect to the 3D case, the volume removal rate $\dot{V}$ by evaporation can be estimated from

$$\dot{V} \approx \frac{A_{x,y} f_{la} (I_{ab} t_R - \phi_{Loss})}{\rho \Delta H_v}, \quad (7.14)$$

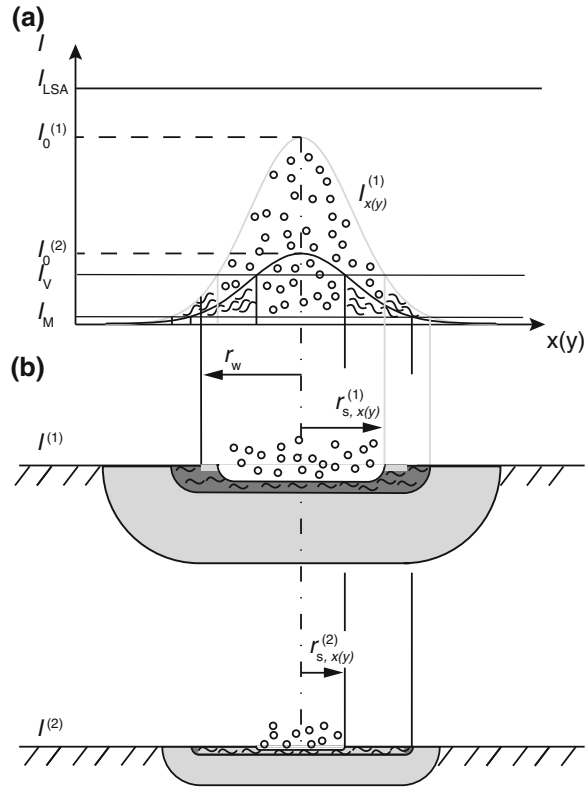
where ϕ_{Loss} describes fluence losses arising in the system due to thermal radiation, heat conduction, convection, etc. When modeling the technological process, one issue is the discrepancy between the high physical ablation rates and low removal rates in real processes. A relatively low $\dot{V}$ mainly originates from the presence of the product $f_{la} t_R$ in Eq. (7.14) which for $f_{la} = 50 \text{ kHz}$ and $t_R \approx t_D = 200 \text{ ns}$ decelerates the milling by two orders of magnitude. Therefore, if machining speed is of particular importance, then the factor $f_{la} t_R$ should be increased. This can be achieved in a plausible manner over playing up the laser repetition rate f_{la} and pulse length t_D . In addition, the time t_R can be increased by applying high intensities providing shorter boiling times t_B and, hence, $t_R \approx t_D$. A more sophisticated approach arises when considering Eqs. (7.12) and (7.14). While the attainment of boiling temperature T_B scales as $I_{ab} \sqrt{t_B}$, the subsequent vaporization rate is $I_{ab} t_R$. This observation suggests the use of modified pulses with high peak intensities at their temporal forefront.

Besides the fact that the ablation is temporally limited to the pulse fraction t_R , it is spatially localized owing to transversal intensity profile of the beam. Assuming the single mode operation in its basic mode TEM_{00} , the intensity distribution along the transversal axis x is described by the Gaussian function $I(x) = I_0 \exp[-2x^2/r_w^2]$ (see Fig. 7.30a). If ablation occurs at a threshold intensity of vaporization $I_V \leq I_0$, then the lateral radius of the ablated chip $r_{s,x(y)}$ can be calculated from

$$r_{s,x(y)} = r_w \sqrt{\ln \sqrt{I_0/I_V}}, \quad (7.15)$$

whereby r_w is given by Eq. (7.4). For a given threshold I_V , the variation in peak intensity I_0 hence tunes the ablated crater radius $r_{s,x(y)}$. While for $I_0 = e^2 I_V \approx 7.4 I_V$ the radius $r_{s,x(y)} = r_w$, for a lower intensity $I_0 = I_V \sqrt{e} \approx 1.65 I_V$ the radius $r_{s,x(y)}$ is only a half r_w . When I_0 approaches the threshold value, i.e., $I_0 \geq I_V$, the

Fig. 7.30 a Allocation of areas of ablation (circles) and melting (waves) in space owing to sample illumination with laser beam exhibiting peak intensities $I_0^{(1)}$ and $I_0^{(2)}$ and Gaussian intensity distribution. I_M , I_V and I_{LSA} denote the threshold intensities for melting, evaporation and plasma generation, respectively. **b** Qualitative description of structural modifications in metals owing to irradiation with pulses of constant duration but different intensity levels. Lower intensities ($I_0^{(2)} > I_V$) yield smaller features



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ablation condition is only fulfilled in a very confined region providing smaller features ($r_{s, x(y)}^{(2)}$ in Fig. 7.30b).

Physically, the evaporation process eventuates when the boiling temperature has been achieved and the delivered energy exceeds the values for the enthalpy of vaporization. Unfortunately this is often mixed with the term of sublimation, which is the transition of a solid into a gaseous phase without passing the fluid one. In the case of evaporation, for a Gaussian beam and long pulses the temperature profile in the matter leads to a melting and subsequent heterogeneous evaporation of atoms or molecules from the surface at the center of the incident beam. This evaporation is, in general, retarded by gas molecules of an ambient atmosphere leading, for some high aspect ratio applications, to limitations in the achievable depth.

In practice, there are a number of different processing techniques. Among them are focusing, projection (mask), interference, microlens array and combined techniques (for an overview see [3]). In the following, only the focusing technique is considered as it is relevant for most applications. It is based on the direct writing of craters on the material's top surface by a focused pulsed radiation while the sample is moved by translations of a xyz -table. Within a fixed table position, in

advance, a laser beam can be rapidly moved over a small sample surface, i.e. beam scanning or beam displacement, by means of galvanoscanner mirrors providing higher speeds but less accuracy and stability. The setup discussed in [94] takes advantage of a Q-switched Nd:YAG laser with $t_D = 100$ ns, $f_{la} = 30$ kHz, $\zeta < 1.2$, $r_W = 9$ μ m providing peak intensities up to 20 GW/cm². While the galvanoscanner window is 40×40 mm² large with a lateral positioning resolution of 1 μ m, the table can be adjusted within a field of 200×200 mm². When scanning the beam in one direction, the train of craters scribes a line into the material; a lateral shift and the scribing a line in the other direction, as shown schematically in Fig. 7.31a, make 2.5D machining possible. Achievement of uniform removal requires the laser traces to overlap in the range of 50–80 %. Nevertheless, application of the bi-directional technique results in a deficient milling due to residual waviness. The latter, however, may almost be avoided by carrying out cross scans (mutual angle 90°, Fig. 7.31b) and wheel scans (mutual angle 15°, Fig. 7.31c). The latter strategy is to be preferred as it suppresses the waviness down to the roughness level. In addition, the optimization of the laser path as well as removal strategies is shown to reduce the thermal load of the material [75]. Concepts from drilling—like percussion, trepanning and helical drilling (see for instance [98])—can be applied to milling within automated part handling of overlapping paths. Furthermore, in order to combine the advantages of fast and precise milling, machining has been proposed to be split into rough and finishing machining [94]. While the first one calls for high laser intensities, the latter takes advantage of higher repetition rates and lower power.

Application of the long pulse lasers to different materials has been frequently reported. Most data are presented for metals and conducting ceramics. Ablation of hard dielectric media like glass, corundum and diamond requires much higher intensities above 10 TW/cm² [57] or an UV radiation for direct bond breaking, both not considered in the current subsection. Experimental employment of a Yb: Fiber laser with $t_D = 20$ μ s and peak intensities up to 770 MW/cm² to stainless steel surface has revealed roughness values $Ra = 10.7$ μ m [35]. Hereby, $d_{min} = 40$ μ m, $h_s = 24$ μ m and $\dot{V} = 2.5$ mm³/s have been achieved. The approach utilizes pulses with an initial spike in power at their forefront. While milling stainless steel by pulsed Nd:YAG laser at 9.5 W average power, $t_D = 10$ μ s, minimal feature sizes of 40 μ m, $h_s = 150$ μ m with $Ra = 2.4$ μ m at

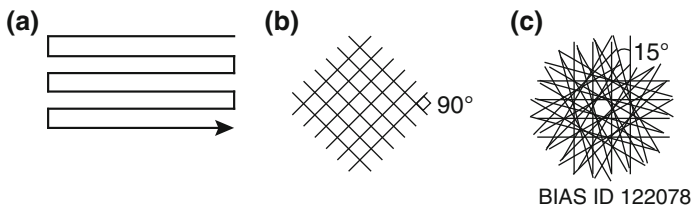


Fig. 7.31 Milling strategies: **a** bi-directional, **b** cross scanning and **c** wheel scanning. According to [94]

scanning speeds $v_{scan} = 400 \text{ mm/s}$ have been demonstrated [75]. Laser drilling of steels has revealed drilling velocities $v_z \sim 10^4 \text{ mm/s}$ above $I_0 = 100 \text{ MW/cm}^2$ [38].

An additional channel of material removal is provided by the positive feedback of material vaporization. The recoil pressure of the evaporated plume may flush off the liquid material by carrying melt droplets from the cavity [16]. Recalling the discussion on average enthalpy, this hydrodynamic expulsion process does not have to supply that large amount of energy and is hence economically efficient. Melting flushing appears at intensities above 10 MW/cm^2 [16] and has been demonstrated to be the dominant mechanism of material removal for pulses longer than $100 \text{ }\mu\text{s}$.

Another ablation process in the long pulse range is due to subsurface explosions [30]. This is relevant for materials with low absorbance coefficient c_λ like non-metallic ceramics yielding $h_\lambda > r_{th}$. In such cases a large volume is heated. Due to the cooling of the surface by dissipative processes, subsurface overheating and explosive ejection of particulate matter may occur. Ablation of WC-10Co with $t_D = 100 \text{ ns}$ has revealed $d_{min} \approx 25 \text{ }\mu\text{s}$, $h_s \approx 15 \text{ }\mu\text{s}$ and a clean surface [26]. Here, the removal mechanism is based on the selective melting and vaporization of cobalt. The removal of the matrix material leads to casting out of WC grains by both the cobalt melt and vapor. Different feedback relations such as absorption of the laser energy in the blow off vapor and expelled liquid while illuminating with long laser pulses have to be considered (see diagram in [16]). Therefore, controlled subtraction requires forefront experimental studies providing a basic understanding of the relationships between different mechanisms and mutual impacts in every particular case which is indispensable for consistent processing.

7.3.2.3 Laser Ablation with Short Pulses

Application of laser pulses ranging from 100 ps to 100 ns is an effective approach of material ablation under condition of suppressed thermal load. Both the recast layer and the HAZ in the matter are reduced owing to the $\sqrt{t_D}$ law. The ablation is called “cold” if the ablated depth h_s is of the order of the affected depth: $h_s \approx \max\{h_\lambda, r_{th}\}$. For many tool materials and UV radiation, this condition is well fulfilled. From the point of the laser source, shortening of pulse length give a positive feedback facilitating higher peak irradiances I_0 . These are highly desired in some applications. Their further rise above plasma ignition threshold I_{LSA} , however, results in plasma shielding of the incident radiation as well as in scattering effects which are highly undesired in intricate milling. Therefore, holding the intensity below the plasma threshold I_{LSA} , short pulse radiation has a disadvantage that the removal rate pro pulse, compared to longer pulses, is limited by fluence factor $I_0 t_R$, visualized in Fig. 7.29, which itself is confined by the product $I_{LSA} t_D$.

Basically, for most metals the processes in nanosecond ablation regime are of photothermal (pyrolytic) nature. The pulse length is longer than the time of electron-phonon scattering, and therefore the mechanism of heterogeneous

vaporization discussed in Sect. 7.3.2.2 may be considered. Advantageous is the better focusability of the UV radiation—due to the lower r_w —providing smaller features. In addition, the light is more efficiently coupled into the material due to low electron plasma screening in this spectral range. In dielectrics with a band gap energy $E_{gap} < 8 \text{ eV}$, the photon energy of the UV radiation may exceed that of bond breaking ($\hbar\omega_{UV} > E_{gap}$) and therefore cause bulk ablation by molecular fragmentation without significant thermal damage owing to the linear absorption of UV radiation. This mechanism is denoted photolytic or photochemical ablation. In fused silica, however, the UV light provided by excimer lasers is neither sufficient to cause photochemical ablation ($\hbar\omega_{UV} < E_{gap} \sim 8 \text{ eV}$) nor are the intensities provided in the ns regime high enough to achieve efficient multi-photon absorption. In this range, the ablation was shown to occur via a two-stage process. First, weak absorption by impurities and defects at the surface or even low absorption in the bulk generates so-called “incubation” centers, which, in the 2nd stage, strongly absorb the incident light ablating the matter at high rates [42].

In practice, UV laser light is delivered by excimer or frequency-tripled Nd:YAG lasers. The latter are favored due to higher exhibited beam quality as well as higher repetition rates compared to that of excimer lasers. In addition to the UV lasers, copper vapor and frequency-doubled Nd:YAG lasers provide nanosecond pulses in the visible range. Usually, both the absorption and, therefore, ablation at the center of the illuminated spot are higher than at its edges, leading to an inherent taper angle of the ablated volume. An add-on used to produce vertical walls is based on the controlled inclination of the incident beam, which is, for instance, utilized in trepanning heads for hole drilling.

Applying excimer laser with $\lambda = 308 \text{ nm}$, $t_D = 33 \text{ ns}$ and $f_{la} = 5 \text{ Hz}$, fused silica has been ablated with $\dot{V} = 3 \times 10^{-3} \text{ mm}^3/\text{s}$ demonstrating feature sizes of $380 \text{ }\mu\text{m}$ [42]. Applying copper vapor laser with $\lambda = 511 \text{ nm}$ and $\lambda = 578 \text{ nm}$, $t_D = 20 \text{ ns}$ and $f_{la} = 10 \text{ kHz}$ to synthetic CVD-diamond and a steel sample MRR of $1.5 \times 10^{-3} \text{ mm}^3/\text{s}$ and $6.5 \times 10^{-3} \text{ mm}^3/\text{s}$, respectively, have been demonstrated [50]. In the case of steel milling, roughness values as low as $Ra = 0.3 \text{ }\mu\text{m}$ have been achieved. Here, a thin sacrificial layer has been applied for surface protection and the angle of the laser incidence has been controlled in order to obtain vertical walls. Machining of the fused silica has failed with this laser system. Application of a Q-switched Nd:YAG with $I_0 \sim 10^8 \text{ W/cm}^2$, $t_D = 63 \text{ ns}$ and $f_{la} = 20 \text{ kHz}$ to tungsten carbide has revealed milling rates $\dot{V} = 5 \times 10^{-2} \text{ mm}^3/\text{s}$ [37].

7.3.2.4 Laser Ablation with Ultrashort Pulses

Machining with ultrashort pulses is an emerging and much more promising approach of tool milling. The time constant which splits pulses into short and ultrashort is given by the material-dependent electron–phonon scattering time t_{e-ph} . In the following, this is assumed to be $t_{e-ph} \approx 10^{-10} \text{ s}$ and, hence, the pulses

$t_D < 100$ ps are called ultrashort. Since the ultrashort pulses pass before the thermal equilibrium is adjusted, the system of electrons and lattice ions can no longer be described by one temperature. At the end of every single pulse, the electrons become overheated and the system is then described by a two-temperature model. In this mode, the HAZ is no longer determined by Eq. (7.6), but can be described according to [3] by

$$r_{th} = 2\sqrt{\kappa_e t_{e-ph}} \quad (7.16)$$

where κ_e is the electron diffusivity. Therefore, the heat affected zone strongly depends on the material properties, and application of femtosecond lasers does not lead to a reduction of the HAZ compared to that under picosecond irradiation. Moreover, pulses below 5 ps were shown to scatter strongly on the cold atmospheric plasma [20]. Therefore, application of femtosecond pulses should be weighed in each particular case. For most applications, picosecond lasers within the 5–10 ps pulse range yield good results. A strong advantage of ultrashort pulses compared to longer pulses is the fact that the ultrafast pulses do not interact with the hot plasma, which starts expanding after the pulse has passed [3]. Therefore, the high intensities which have been blocked by plasma ignition in the longer pulse processing can now be utilized for the efficient ablation of dielectric materials. Here, intensities for optical breakdown can be achieved in the femtosecond range whereby the dielectric material is ionized into a very dense and hot plasma.

Physically, different processes are responsible for ablation in ultrafast regime. In metals, in addition to surface evaporation, thermal mechanisms at overcritical temperatures are involved. Among them are processes such as explosive (homogeneous) boiling and phase separation by spinodal decomposition (for an overview see [57]). In dielectrics, multi-photon and avalanche ionization may lead to an optical breakdown, changing the optical properties from transparent to metal-like. Intense radiation then efficiently couples to the electron plasma and initiates strong photoemission of electrons. As a consequence, an electrostatic repulsion between positive lattice ions may overcome the binding forces resulting in a Coulomb explosion (for an overview see [3]).

For ultrafast processing, mode-locked Nd:YAG (Nd:YVO₄) lasers with pulses in the picosecond range or femtosecond Ti:Sapphire laser systems have been utilized. Due to the high costs and complex design of Ti:Sapphire lasers, picosecond laser sources have mostly obtained industrial relevance.

In the following, selected experimental data are presented. By milling steels with mode-locked laser with frequency-tripled Nd:YVO₄ laser with $\lambda = 355$ nm, $t_D < 15$ ps and $f_{la} = 30$ kHz, a surface roughness of $Ra = 0.3$ μm has been achieved [50]. In alumina roughness values of $Ra = 0.25$ μm have been achieved [48]. Employment of a laser source with $\lambda = 355$ nm, $t_D = 8$ ps and $f_{la} = 10$ kHz to fused silica has revealed $Ra = 0.3$ μm [49]. Applying the radiation of Ti:Sapphire laser with $\lambda = 800$ nm, $t_D = 150$ fs and $f_{la} = 1$ kHz to WC-10Co and industrial diamond, ablation rates of 2×10^{-4} mm³/s and 2×10^{-5} mm³/s, respectively, have been obtained [26]. Illuminating sapphire with the radiation of Ti:Sapphire laser with

$\lambda = 780 \text{ nm}$, $t_D = 164 \text{ fs}$ and $f_{la} = 1 \text{ kHz}$, ablation rates of $0.3 \times 10^{-5} \text{ mm}^3/\text{s}$ and $1.5 \times 10^{-5} \text{ mm}^3/\text{s}$ for gentle and strong ablation regimes, respectively, have been demonstrated [78]. Here, the minimum structure diameter $d_{min} = 4 \text{ }\mu\text{m}$ and depth $h_s = 0.22 \text{ }\mu\text{m}$ have been reported, demonstrating the state of the art in producing intricate shapes by laser ablation in hard-to-machine materials.

7.3.3 Laser-Induced Chemical Machining

Salar Mehrafsun

7.3.3.1 Process Fundamentals

Laser-induced chemical machining is based on the laser-induced thermal activation of a heterogeneous chemical reaction between an ambient and solid medium. This process can be used for subtractive methods like surface modifications and the milling of micro forming tools [3].

For the processing of metals, wavelength-independent thermal processes can be efficiently applied. The influence of laser radiation causes chemical reactions on the metal surface only, which are associated with material removal (for laser-induced chemical machining) or deposition, depending on the composition of the ambient medium. Due to the high concentration of reactants thermochemical machining is particularly efficient in reactive fluids like phosphoric or sulfuric acid. In contrast with laser ablation with laser power densities up to the GW/cm^2 range, material processing with laser-induced chemical machining already starts at laser power densities of $0.1 \text{ MW}/\text{cm}^2$ [96]. Due to the low laser power densities, thermal and mechanical loading of the material is avoided. Also a redeposition of evaporated and/or molten material inside and outside the laser irradiated area (see Fig. 7.32), which may lower the quality of the machined structure, does not occur [40].

Selective material removal is possible in all metals which are protected against corrosion by a thin natural passivation layer at room temperature. The passivation layer leads to many metals having a negligible background etching rate of $<10^{-8} \text{ }\mu\text{m}/\text{s}$ without laser radiation [69].

The material removal is a result of laser-induced thermochemical reactions between an etchant and metal atoms on the surface of the workpiece, beginning at low laser powers of $<1 \text{ W}$ [69]. The temperature distribution at the surface of the workpiece is an important factor in material removal. Investigations with continuous wave (cw) Ar-ion ($\lambda = 514 \text{ nm}$) and cw Nd:YAG laser sources ($\lambda = 1,064 \text{ nm}$) show that the passive layer is not affected up to a temperature of $90 \text{ }^\circ\text{C}$, and that equal temperatures lead to equal normalized removal rates for different spot diameters and laser wavelengths [96]. Thereby only the laser wavelength has an indirect influence on material removal due to differences in the absorptivity of the materials. Therefore, the reaction leading to material removal only depends on the temperature of the material surface and not on the absorbed

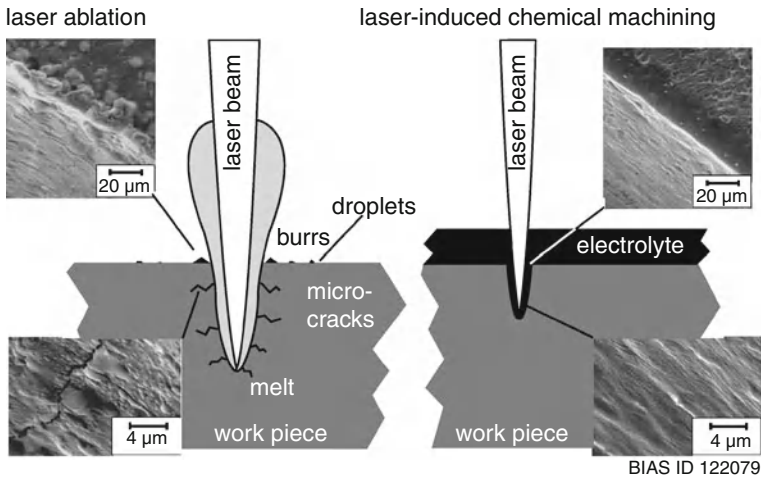


Fig. 7.32 Comparison of processing quality in (*left*) laser ablation and (*right*) laser-induced chemical machining

laser wavelength [96]. For metals with a dense natural oxide layer, thermal conductivity reflects the achievable processing speeds. For NiTi alloys processing rates of up to $8 \times 10^4 \mu\text{m}^3/\text{s}$ with laser powers of only 7 W can be achieved [94]. The achieved temperature at surface level is crucial to predict the material removal and achieved contour. For laser beams arranged perpendicularly to the surface of the workpiece the etched depth h can be calculated in dependence on the laser power P_l and feed rate v by the semi-empirical equation:

$$h = \frac{De^{-\frac{P_a}{P_l}}}{v + v_0} \quad (7.17)$$

where D is a pre-exponential factor with the units of a diffusion coefficient (m^2/s), P_a the laser activation power and v_0 a coefficient considers that the maximum etching depth, even for a feed rate of zero, is limited. These coefficients do not depend on the processing conditions but on the material properties [95]. Investigations show that the calculated etch depths correlate well with experimental results (see Fig. 7.33). For laser powers of 12 W etch depths of up to 380 μm for steel X110 and 150 μm for Stellite 21 can be achieved [95].

Physically the process can be described by thermochemical phenomena leading to etching reactions. In aqueous solutions the metallic workpiece behaves like an electrode. A potential difference can be observed due to the formation of an electrolytic double layer at the phase boundary. Metal ions which enter the solution leave the electrode in an electron surplus. According to Helmholtz, metal cations near the electrode are attracted and it comes to a formation of a simple double layer (Fig. 7.34). This is also called a rigid Helmholtz double layer, in contrast with the thereafter forming diffuse Nernst layer, which is followed by the

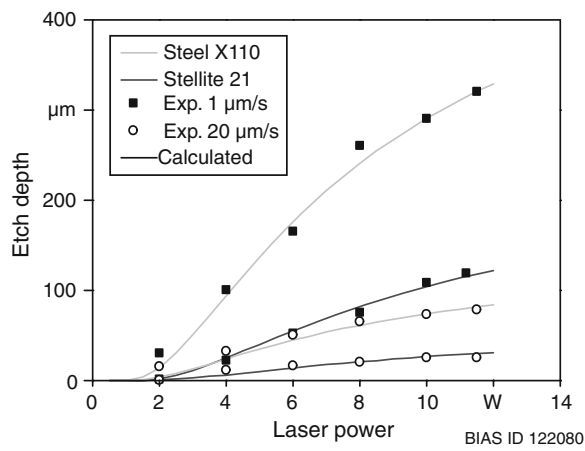


Fig. 7.33 Calculated and measured etch depths versus laser power using different feed rates for Stellite 21 and steel X110 [95]

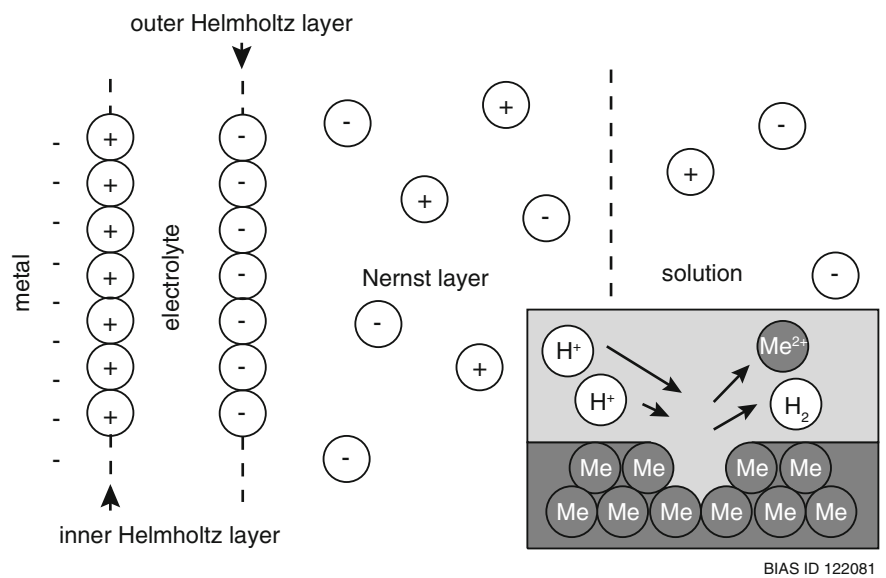
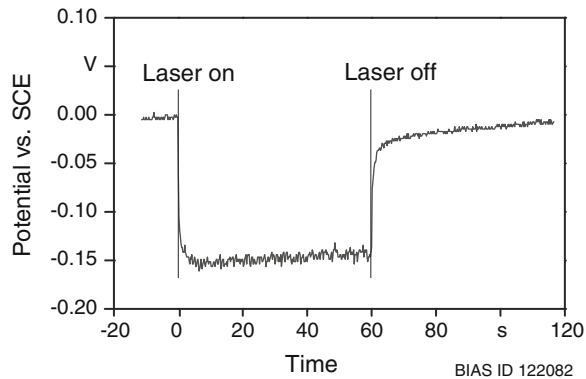


Fig. 7.34 Schematic illustration of the electro-chemical double layer on an electrode and chemical reaction between the metallic surface and the reactive wet ambient medium

solution. Consequently the charge with increasing distance to the electrode becomes weaker and the concentration of reactants increases [11].

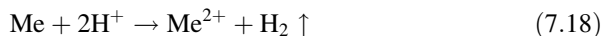
The time-resolved measurements of the electrochemical potential ψ_{el} against an electrochemical reference electrode (saturated calomel electrode SCE) show that the passivation layer is locally removed, followed by a subsequent dissolution of

Fig. 7.35 Temporal evolution of the potential for an NiTi alloy switching the laser on/off [94]



the metal under the formation of hydrogen and water soluble metallic salts. This is indicated by a sudden drop in the voltage down to -0.15 V caused by the thermal influence of the laser beam (see Fig. 7.35). After the end of laser irradiation an immediate interruption of the etching reaction is detected and a sudden increase in voltage due to the repassivation of the metal surface, which is essential for high quality machining [94].

Due to the chemical dissolution of the material (see Fig. 7.34 right), according to Eq. (7.18) the activity of the products (hydrogen and metal ions) increases, at the point where the activity of the educts (protons and metal) decreases [96].



According to Nernst the decrease of proton activity causes a decrease of the potential, while the increase in metal ion activity results in a positive change to the electrochemical potential ψ_{el} . Therefore the chemical material removal within the reactive fluid is driven by the temperature-dependent proton activity of the redox reaction, and is mainly responsible for the formation of the electrochemical potential [96]. This effect can be used for the effective and precise processing of a range of different metallic materials (see Sect. 7.3.3.3). By applying different external application-oriented electrical potentials it is possible to machine workpieces with high shape quality and contour accuracy (see Sect. 7.3.3.2). The laser supports the kinematics of the metal dissolution without the melting effects (see Fig. 7.32). The electrochemical potential ψ_{el} of the workpiece not only depends on the laser power-dependent temperature T but on the chemical activity i_{ch} of the dissolving metal ions. According to Nernst the electrochemical potential is proportional to the temperature, and therefore to the laser power applied [96].

Another fundamental influence on the processing speed beside the thermal activation of chemical reactions is the mass transport limitation within the liquid ambient medium. As previously mentioned, a non-moving layer is formed in heterogeneous reactions in front of the solid surface (Helmholtz and Nernst layer). The reactants must pass through this layer to react with the solid surface (see

Fig. 7.34). The reactants can only get through this layer by diffusion. The Nernst layer is depleted of reactants by their reaction with the solid surface. The concentration of the reactants near the solid surface at the outer Helmholtz layer may decrease to a level where not enough educts can be delivered [11]. Investigations show that for small removal depths the removal rate fluctuates around a constant value. For larger depths a monotone decrease with a gradient of $m = -1$ can be observed, i.e. the reaction rate is inversely proportional to the removal depth [96]. For a heterogeneous chemical reaction with mass transport limitation the reaction rate is a function of the diffusion coefficient D and the thickness of the Nernst layer δ . Therefore, the processing speed in laser-induced thermochemical machining is diffusion limited [96]. For the precise machining of e.g. micro metal forming tools the influence of laser power, electrochemical potential and mass transport limitations are important factors, which have to be taken into account.

7.3.3.2 Process Method

In practice, a significant disadvantage of laser-induced chemical machining is the low process speed (material removal rate = 1.6×10^{-5} mm³/s), which is an obstacle for use in industrial applications. An increase in laser power leads to higher material removal rates but can also cause molten surfaces and hence a deterioration in processing quality. Nevertheless the laser irradiated surfaces can reach higher temperatures than the boiling temperature of the reactive fluid, leading to the formation of gas bubbles which may slow down or interrupt the etching reaction [22]. Resulting in a decrease in material removal and also in processing quality. Since etching processes are transport limited, a further significant factor is the increasing metal ions density in the processing zone which also leads to a decrease in material removal [69]. Thus a fast exchange of the reactants is necessary to avoid saturation effects in the etching reaction.

One approach is the application of a reactive wet jet-stream. With the help of a jet nozzle an efficient and fast exchange of the reactants can be realized. It leads to increased removal rates one order of magnitude (material removal rate = 1.6×10^{-4} mm³/s), due to the intensive mass transport by direct injection of the etchant into the laser irradiated area [94]. Additionally, the etchant jet-stream cools the workpiece surface. This leads to a suppressed boiling of the etchant and results in an improvement in the processing quality and shape fidelity [22]. The main components of this approach are a fiber-laser, a xyz-linear stage and a liquid-phase etching cell as shown in Fig. 7.36. The continuous wave (cw) fiber-laser source is operating at a wavelength of 1,080 nm. The focused laser beam is guided coaxially to the etchant jet-stream through the nozzle onto the workpiece surface. The workpiece is fixed in a basin, which is mounted on the computer-controlled xyz-linear stage. It allows feed rate v control of the workpiece and adjustment of the workpiece with respect to focal position. An adjustable pump enables flow rates of the etchant jet-stream between 2 and 20 m/s.

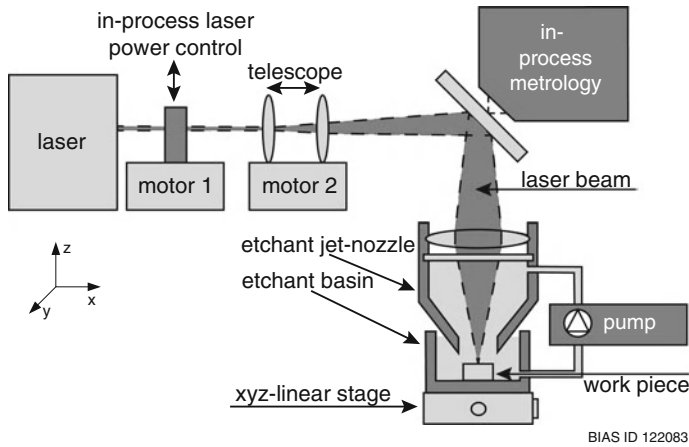


Fig. 7.36 Setup for laser-induced chemical machining with an etchant jet-nozzle and functional schematic of liquid phase etching cell

Investigations show that with the right flow rate setting there is an increase in processing quality [64].

Additionally a three-electrode configuration is integrated into the etching cell to perform the electrochemical enhancement of the reaction if required. For an automatic operation of the system a user interface is integrated. With this user interface it is possible to set and control the processing parameters to machine a desired structure. A quality control system optimizes the process stability and the manufacturing accuracy, enabling flexible and precise machining of hard and geometrically-complex micro tools [64].

Material removal is strongly influenced by four main process variables namely laser power P_L , feed rate of the workpiece v , flow rate of the etchant u and focus diameter d of the laser beam spot. The intensity profile of the fiber-laser is close to a Gaussian bell curve. Combining several bell curves leads to the desired shape. Laser path planning calculates the number, position and Gaussian intensity profiles for a sequence of overlapping laser removal paths (cavities), as shown in Fig. 7.37. The inverse process model calculates the required process parameters according to the desired removal path. The calculated process variables are fed into the in-process control block, individually observed and controlled by the graphic user interface. The cross-section of a desired individual cavity can be approximated by Gaussian curves. By changing both Gaussian form parameters, amplitude and standard deviation (depth and width of the cavity), the cross-section of individual removal paths can be varied and their superposition can be matched to a desired contour [64]. An appropriate combination of process parameters allows successful material removal and contouring of e.g. micro metal forming tools.

The chemical etching reaction can be further influenced by an external electric field, between the metal workpiece and an additional electrode. This is possible due to the ionic nature of the reactants [94]. In contrast with laser-assisted

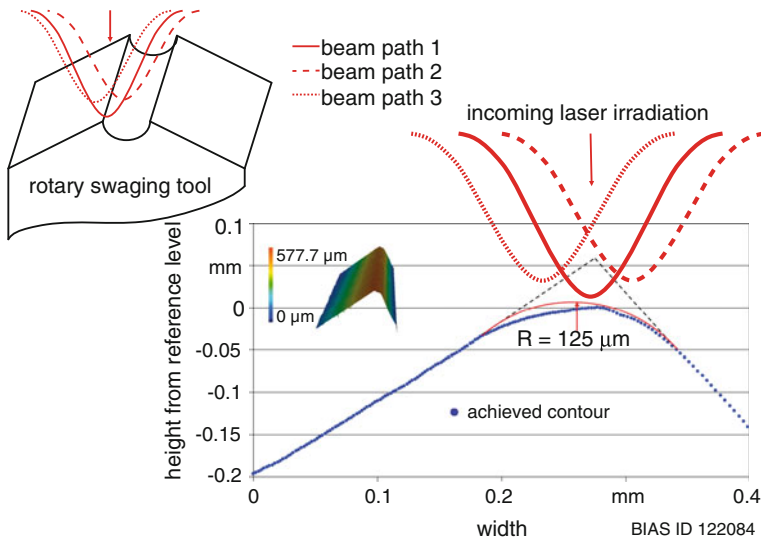


Fig. 7.37 (left) Schematic representation of sequential removal paths for a rotary swaging tool and (right) manufactured edge radius

electrochemical machining, where material removal is electrochemically dominated (more than 100 V) [22], the applied voltage in laser-induced chemical machining is very low (a few volts) [94]. Laser-induced currents are observed due to the thermal activation of electrochemical etching reactions and/or laser-induced breakthrough of the passivation layer. Depending on the polarity of the applied electric field, the increase in etching rate (reduction of the passivation layer towards the cathodic potentials) can double ($\text{MRR} = 3.2 \times 10^{-4} \text{ mm}^3/\text{s}$), or an improvement of the processing quality with e.g. R_a down to 0.1 μm (stronger passivation) can be realized. Even materials which are not protected by a thin native oxide layer such as hardened tool steel can be electrochemically passivated and therefore protected against corrosion outside the laser irradiated area by applying an external electric field [94]. Thus, combining the influence of five independent process parameters (laser power, etchant flow rate, feed rate of the workpiece, focus diameter and electric field), optimum conditions for obtaining desired shape contours can be found.

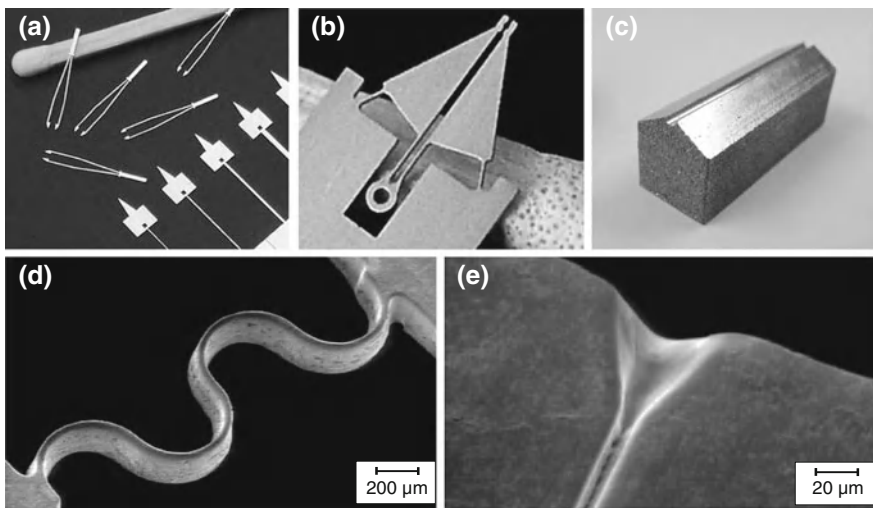
7.3.3.3 Achievable Structures

With laser-induced chemical machining it is possible to generate complex geometries and micro structured surfaces. It allows the machining of a number of different materials and alloys. In particular, titanium, nickel and chromium-based alloys, stainless steel as well as high performance and special materials such as Stellite alloys are highly processable. Achievable structures are not limited in maximum

size because they are successively generated. Cavities can be machined by moving the workpiece perpendicular to the laser beam. The width and depth of the cavities are influenced by the temperature distribution on the surface and the duration of the temperature rise owing to the thermal activation of chemical reactions [96]. Depending on the nature of the material, minimal cavity widths down to 10 μm and cavity depths up to 380 μm by single scanning can be realized [95]. Therefore, the shape of the cavity reflects the effective heat conductivity of the machined material and the intensity distribution of the incident laser beam [94]. In addition to its capability in micro forming tool machining, laser-induced chemical machining is also suitable for the manufacture of micro functional elements like micro grippers, micro springs and micro channels, as shown in Fig. 7.38.

With minimum structure sizes of 10 μm , cutting angles of 3° , edge radii of 2 μm and a mean roughness R_a in the range of 0.1 and 0.3 μm , achievable micro structures are in the field of competing processes such as micro milling, EDM or ECM. Because of the lateral heat diffusion from the laser processing zone, with increasing laser power the depth and width of the cavities simultaneously increase. Thus, by single scanning of the cavity, aspect ratios (etched depth to width) of 3 can be realized. However, by multiple scanning of the laser beam along the same cavity, higher aspect ratios (up to 10) can be realized [69]. The feed rate of the workpiece can be up to 100 $\mu\text{m}/\text{s}$ and material removal rates of up to $3.2 \times 10^{-4} \text{ mm}^3/\text{s}$ can be achieved [94].

With the laser path planning algorithm, the number, position and Gaussian intensity profiles for overlapping laser removal paths can be calculated. As shown



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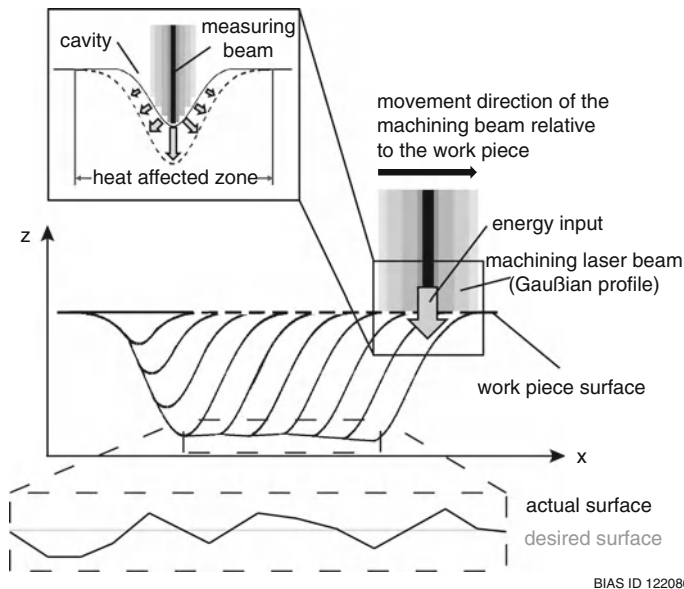
Fig. 7.38 Laser-induced chemical machining manufactured micro grippers made of nitinol (a, b), contoured micro rotary swaging tool (c), micro spring made of nickel-titanium (d) and flexible shaped cavity in nickel (e)

in Fig. 7.37, the edges of a micro rotary swaging tool can be contoured with a defined radius. For the precise material processing of e.g. micro rotary swaging or micro deep drawing tools a quality control system which includes an automated laser path planning model is required. Additionally an in-process metrology via interferometry for a depth control of the cavities helps improve the processing quality and shape fidelity.

7.3.3.4 In-situ Process Monitoring

Influencing factors such as minimal fluctuations in laser beam intensity or etchant flow rate and material inhomogeneities affect the rate of material removal. The result is a deviation in the desired width and depth of the cavity. Knowledge of the actual geometry allows the control of the machining process and thus meets the requirements of high accuracy and short processing times. For laser removal processes with limited spatial accessibility, coaxial optical measuring techniques offer suitable solutions for the given measuring tasks.

Here, interferometry represents a well-established and stable measuring technique with high resolution. A laser beam is used to measure the current leveling of the cavity depth as shown in Fig. 7.39. The measuring laser beam is guided coaxially to the machining laser beam, and the jet-stream onto the workpiece surface. Depending on the measuring beam wavelength used, the measurement



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Fig. 7.39 (above) Arrangement of the measuring beam within the processing beam, (below) illustration of the machined cavities in laser-induced chemical machining

accuracy reaches the nanometer range. But with increased measurement accuracy the axial resolution decreases. In this case multiple wavelength measurement systems have to be implemented to ensure successful and precise measurement.

Having the necessary information about the depth of the cavity, permanent adjustment control of the process parameters is possible. This will lead to the desired depth of the cavity and thus to the desired contour. By applying in-process metrology to Laser-induced Chemical Machining, high surface quality, burr-free cut edges and high shape quality can be achieved. Thus, the rejection of parts and the processing time for good parts can be reduced.

7.4 Cutting and Abrasive Processes

Ekkard Brinksmeier, Olthmann Riemer, Sven Twardy and Christian Robert

7.4.1 Machining Processes and Machine Tools

The machining of hard and wear resistant materials—like tool steels or cemented carbides—at a micro scale is even more challenging than at macro scale, because machining accuracy cannot be scaled down linearly and cutting tools are becoming much more fragile when the tip sizes are reduced. Therefore, to achieve a high quality in manufacturing, very precise and stiff machine tools in combination with suitable processes have to be used. But without doubt, micro manufacturing processes based on mechanical material removal exhibit a great potential for the flexible machining of micro forming tools in single piece production and for small batches. Nevertheless, there is a tradeoff between surface quality and material removal, like that shown in Fig. 7.40. For this reason, processes often have to be combined in suitable process chains to achieve high quality products at reasonable cost. In the following sections the general characteristics of mechanical micro machining processes will be introduced, and the suitability for micro mold making will be discussed in detail.

In general, all mechanical micro machining processes require machining systems which have to fulfill specific requirements regarding the machine tool itself, the environment, the subsystems and the data management, i.e. the CAD-CAM chain. Figure 7.41 shows a CNC precision micro machining center and essential subsystems. For all precision machining operations four major areas have to be taken into account, as described in the following:

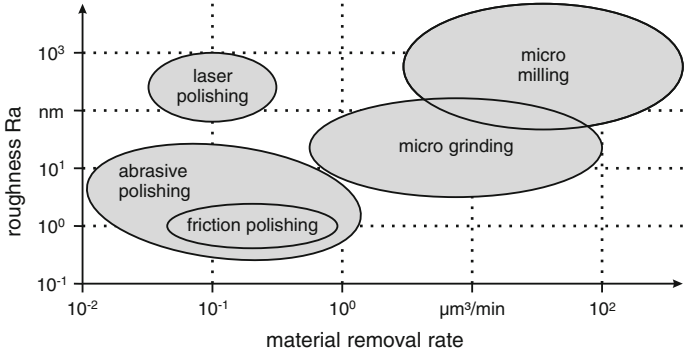


Fig. 7.40 Machining processes for micro forming tools categorized by typical material removal rates and achievable surface roughness

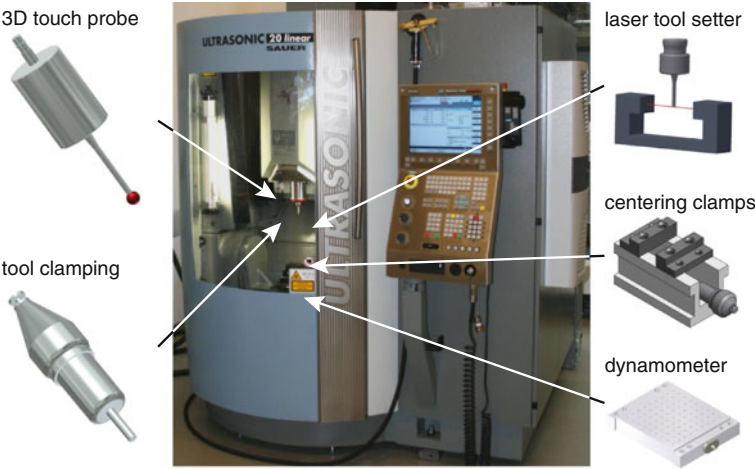


Fig. 7.41 Precision micro machining center (DMG Sauer Ultrasonic 20 linear) and subsystems

7.4.1.1 Machine Tool

The applied machine tools should be built from state of the art technologies by means of linear direct drives, hydrostatic guideways and glass scales for position measurement and assembled on a frame with a high damping rate. For small cutting tools a high frequency spindle ($n > 20,000$ rpm) is also compulsory in order to achieve sufficient cutting speeds.

7.4.1.2 Environment

Stable environmental conditions are essential to achieve high accuracy. For this reason temperature and humidity should be controlled, and external influences (e.g. sunlight, air flow and vibrations) have to be isolated from the machine tool.

7.4.1.3 Subsystems

For high precision machining, a precise machine tool by itself is not sufficient, but also high quality subsystems are necessary. For the lowest tool run out ($<3 \mu\text{m}$) a precise tool clamping system with high balance quality (G 2.5 or better) should be used. For accurate determination of the tool's center point and the radius at spindle speed a laser tool setter is recommended. The workpiece should be clamped as near as possible to the center of the rotational axes to minimize angular errors. The exact position of the workpiece should be determined by a 3D-touch probe. For the measurement of cutting forces a piezo-dynamometer can be applied.

7.4.1.4 CNC control

The CNC control must allow very short block processing times to enable accurate high speed machining. Error compensation for guideways and thermal expansion is useful to improve machining accuracy. An adapted interface to the CAD/CAM system is essential for loss-free data transfer.

7.4.2 Cutting Processes

All material removal processes for which the number of cutting edges, the macro geometry of the tool and the tool orientation are known, are defined as cutting processes. Through these processes workpiece material is removed mechanically in the form of chips. The most commonly used cutting processes in mold making are turning (for rotational symmetric parts) and milling. Milling in particular is widely used because of its high flexibility (non symmetrical parts and freeform surfaces), removal rate and surface quality. If the milling process is scaled down to a certain limit it is called micro milling. Micro machining has two popular definitions:

- structural dimensions smaller 1 mm by Masuzawa and Tönshoff [61]
- tool engagement less than 1 mm by Dornfeld et al. [25].

Molds for micro forming processes are usually larger than 1 mm, but contain features in the micrometer range. For this reason in most cases mold making is a combination of conventional milling and micro milling, because not all cutting tools applied are smaller than 1 mm and not all machined features are smaller than 1 mm.

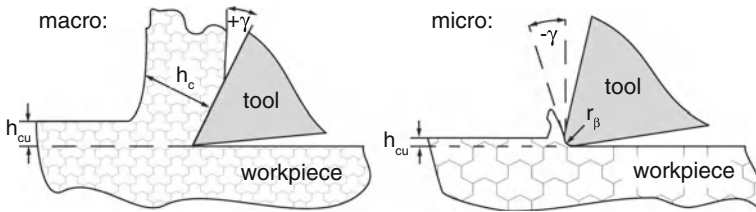


Fig. 7.42 Size effects in down scaling of cutting processes: conventional macro-scale (*left*) and micro-scale cutting (*right*) after [2]

But the milling of hardened steel with small cutting tools even with diameters in the range of a few millimeters is quite challenging, because there are two dominating size effects when the milling process is scaled down (cf. Fig. 7.42):

- the workpiece grain size is in the same order-of-magnitude as the uncut chip thickness h_{cu}
 - material behavior becomes anisotropic
- the uncut chip thickness h_{cu} is in the same order-of-magnitude as the cutting edge radius r_β
 - effective rake angle γ becomes negative.

If the grain size of the workpiece material is in the same order-of-magnitude as the uncut chip thickness h , the material will behave anisotropically, because the interaction between the workpiece and cutting edge is limited to single grains with different crystal orientations or alloying. This effect can cause strong variations in the cutting conditions and chip loads, resulting in poor surface quality or chatter. For this reason fine grained tool steels with dispersed carbides are recommended for micro milling.

If the uncut chip thickness h is in the same order-of-magnitude as the cutting edge radius r_β , the theory of minimum chip thickness becomes very important. For micro milling of hardened tool steel, Aramcharoen and Mativenga have shown that the minimum chip thickness $h_{cu,min}$ correlates approximately to the value of the cutting edge radius, which is typically $0.5 \mu m < r_\beta < 2.5 \mu m$, in micro milling, depending on the cutting material and coating. This effect is shown in Fig. 7.43 [2].

According to the theory of minimum chip thickness three cases can be distinguished:

- | | |
|---------------------------|--|
| $h_{cu} < h_{cu,min}$ | No material will be removed in the form of chips. Only elastic and plastic deformation occurs, which is called plowing. For $h_{cu} \ll h_{cu,min}$ no plastic deformation occurs and material will recover completely |
| $h_{cu} \cong h_{cu,min}$ | Material is partly removed in the form of chips, but plowing can still occur |
| $h_{cu} > h_{cu,min}$ | Material is completely removed in the form of chips |

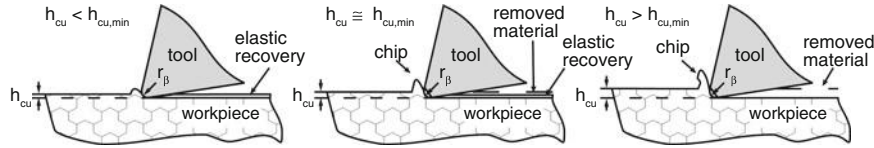


Fig. 7.43 Chip formation relative to minimum chip thickness $h_{cu,min}$ in micro scale machining after [2]

Cutting tests have shown that the best surface quality in ball end milling of hardened tool steel can be obtained if the uncut chip thickness h_{cu} , is in the range of the minimum chip thickness $h_{cu,min}$. If the uncut chip thickness is too low, plowing effects (elastoplastic material response) occur due to the negative rake angle. On the other hand, if the uncut chip thickness is relatively high, regular feed marks can be observed, leading to a rough surface profile, which is dominated by the kinematics of the cut (cf. Fig. 7.44).

Despite these limitations, micro milling is an appropriate technology for the machining of tool steels with a hardness of up to 70 HRC, if performed on a suitable machine tool, applying appropriate cutting tools and well selected cutting parameters.

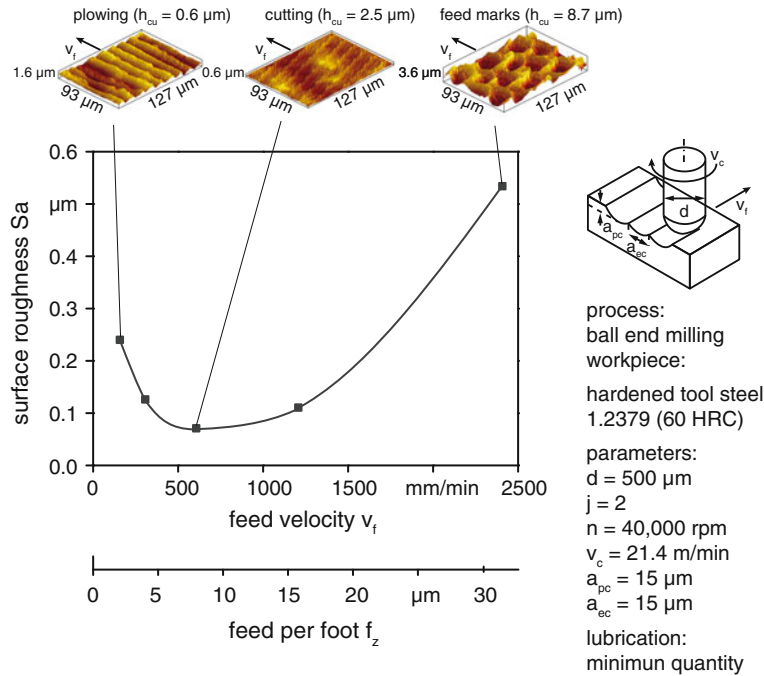
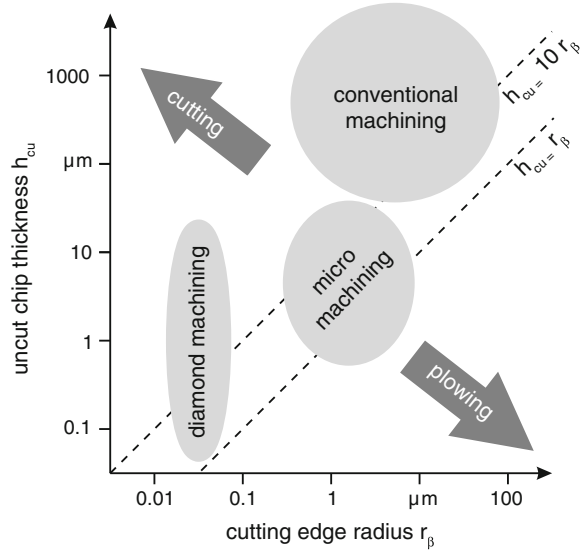


Fig. 7.44 Surface roughness S_a versus feed velocity v_f and feed per tooth f_z in micro ball end milling of hardened tool steel

Fig. 7.45 Typical uncut chip thickness versus cutting edge radius for different cutting processes



A further reduction in the uncut chip thickness without plowing is only possible in diamond machining, because of the very sharp cutting edge ($20 \text{ nm} < r_\beta < 50 \text{ nm}$) (cf. Fig. 7.45). Unfortunately diamonds cannot be applied as cutting tool materials for the machining of iron based metals like steel, because of the excessive wear rate due to the high affinity for carbon atoms of the iron.

7.4.2.1 Cutting Tools

In conventional machining many different cutting tool materials, like High Speed Steel (HSS), tungsten carbide, ceramics, cermets, CBN, diamond, etc., have been adapted for individual machining tasks. In micro milling of hardened steel only two common cutting tool materials are used, i.e. tungsten carbide and CBN. Typical cutting tools are shown in Fig. 7.46. Micro cutting tools are shaped by grinding on high precision multi-axes grinding machines. The cutting edge radius of uncoated tools is usually smaller than $1 \text{ } \mu\text{m}$ and for coated tools between 1 and $5 \text{ } \mu\text{m}$. The number of cutting edges j is usually two for micro milling tools, and in rare cases four. High quality tool manufacturers provide measurement protocols for all delivered cutting tools showing nominal/actual diameters and run out errors between the shank and cutting edges.

7.4.2.2 Material Removal Rate

The material removal rate Q_w is defined as a product of the depth of cut, width of cut and feed velocity:

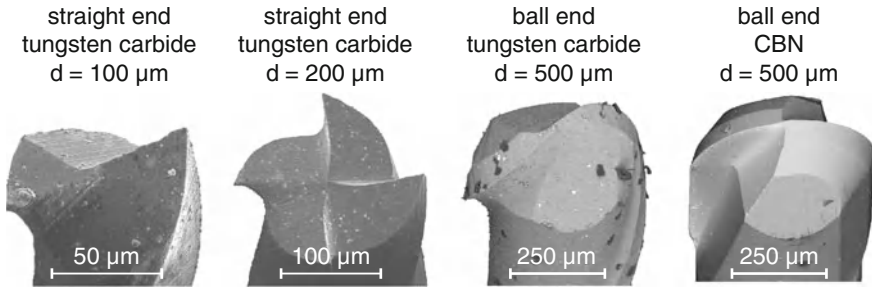


Fig. 7.46 SEM pictures of state of the art micro milling tools

$$Q_w = a_{pc} \cdot a_{ec} \cdot v_f \quad (7.19)$$

In micro milling the optimal removal rate is mainly dependent on the combination of cutting tool and workpiece material. For each combination there is an optimum cutting speed v_c . All other parameters are dominated by the cutting tool diameter. For this reason and for stability aspects cutting tools should always be chosen as large as possible. Figure 7.47 shows typical material removal rates versus cutting tool diameters in the ball end milling of hardened tool steels by applying coated tungsten carbide tools. In straight end milling the material removal rate is slightly increased, because a greater width of cut can be chosen.

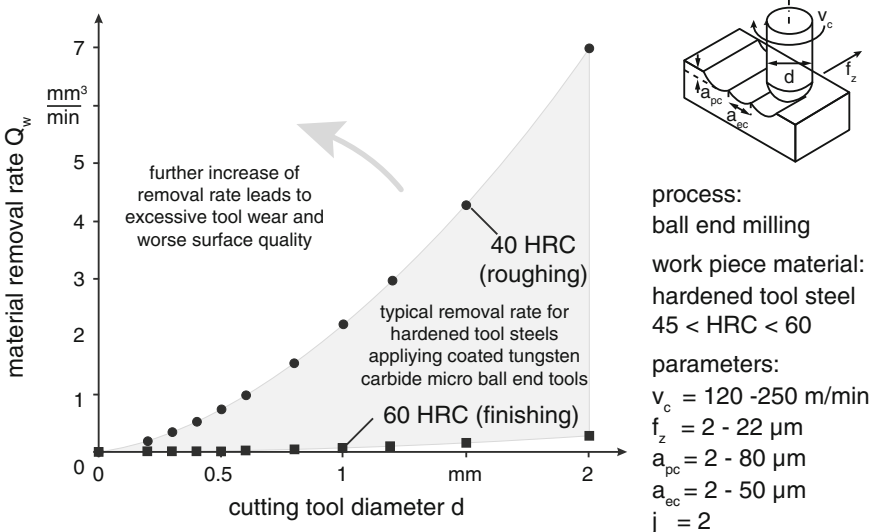


Fig. 7.47 Material removal rate versus cutting tool diameter in micro ball end milling of hardened tool steel, based on cutting parameter suggestions of two different cutting tool manufacturers

7.4.2.3 Shape Accuracy and Surface Quality

The achievable shape accuracy mainly depends on the positioning accuracy of the applied machine tool. For precision machining in terms of micro milling, shape accuracies of some micron can be achieved. For higher precision ultra precision machine tools have to be applied. Besides positioning accuracy, tool workpiece interaction and the resulting tool deflection are important factors limiting shape accuracy, for this reason cutting forces should be minimized. Achievable aspect ratios of structures are often only limited by tool size. Today, cutting tools for milling and drilling are commercially available down to a diameter of 50 μm . Smaller tools have also been manufactured and tested by research institutions.

After form accuracy, surface quality is the major quality criteria in mold making. The surface finish in the forming zone in particular is a matter of concern, because this area has a strong influence on the quality of the formed part. Rough forming tools can damage blanks and induce defects during the forming process. But not always does a smooth surface perform best. Many different studies have shown that well-defined micro structures can improve the performance of forming tools.

Nowadays, surfaces are mainly inspected by optical systems like white light interferometers, but also tactile profilers are still widely used. Regardless of the measuring technique, surface parameters were calculated from recorded raw data to evaluate surface qualities. In micro ball end milling surface qualities down to $S_a = 50 \text{ nm}$ can be achieved on hardened tool steel surfaces under ideal cutting conditions.

By the adaption of cutting conditions in terms of tool diameter, feed per tooth and width of cut, various surface topographies with different tribological conditions can be achieved, but only if the minimum chip thickness is exceeded. If not plowing occurs, and the surface topography becomes undefined. Figure 7.48 shows how the variation in width of cut (i.e. line pitch) affects the surface topography expressed by roughness values and the Abbott-curve.

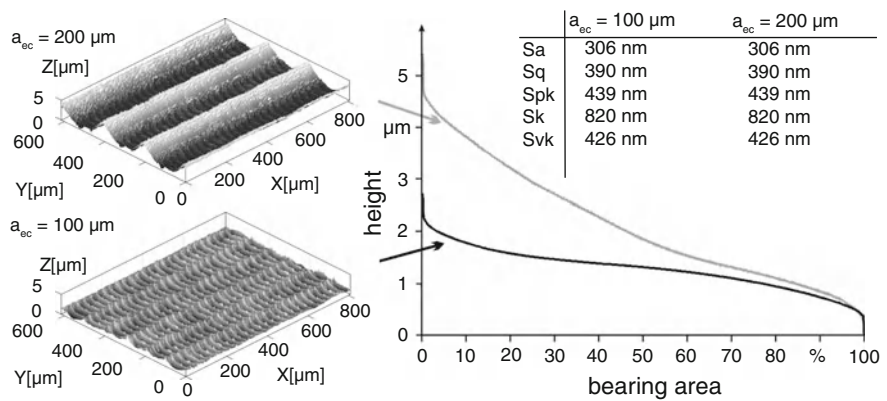


Fig. 7.48 Surface topography and Abbott-curve of micro ball end milled surfaces

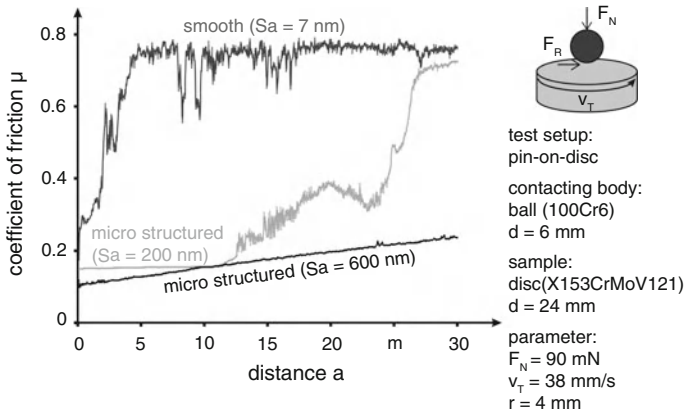


Fig. 7.49 Coefficient of friction over sliding distance of pin-on-disc test of micro structured surfaces [104]

7.4.2.4 Tribological Behavior

In most technical applications very smooth surfaces with a low roughness are desired. But in tribological systems (e.g. forming processes like deep drawing) smooth surfaces do not always perform optimally [9]. Especially in dry contact, smooth surfaces tend to wear out very fast. Investigations on the influence of micro structures on the tribological conditions in dry contact show that micro structures can reduce the coefficient of friction and wear rate, compared to smooth surfaces. Figure 7.49 presents the results of pin-on-disc tests. In these tests only surface topography was varied, all other parameters were kept constant. Micro structured surfaces were machined by micro ball end milling in a radial direction and compared to a polished reference sample. It is obvious, that the polished surface immediately starts to wear out and hence the coefficient of friction increases due to debris in the contact zone. Micro structured surfaces can enclose wear particles much better than smooth surfaces, and exhibit a much more constant tribological behavior [104].

7.4.2.5 Cutting Forces

Cutting forces are a major parameter for the evaluation of machining processes, especially to analyze the influence of different cutting tools, workpiece materials, cooling lubricants and machining parameters on the work result or to identify chatter. Furthermore, high frequency recording of cutting forces by a piezo-dynamometer can deliver helpful information about the tool–workpiece interaction. A typical force plot is shown in Fig. 7.50. It demonstrates that there is a variation in feed force F_y according to the tool angle, which can be explained by

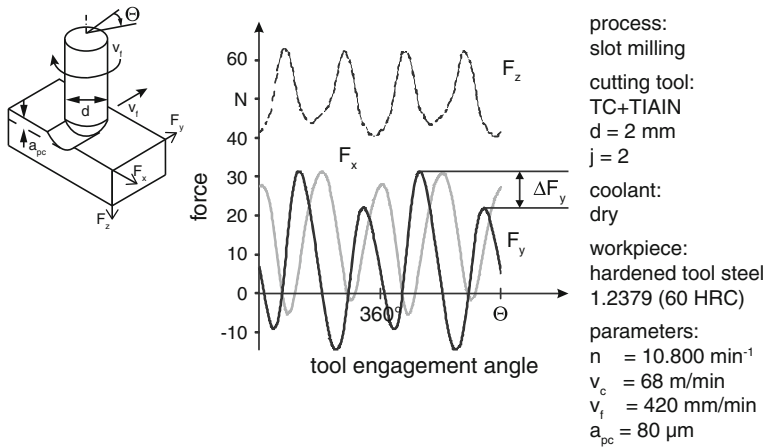


Fig. 7.50 High resolution measurement of cutting forces in micro ball end milling of hardened tool steel

the run out error of the applied cutting tool. The run out error causes a variation in uncut chip thickness for the individual cutting edges. This effect results in higher loads for one cutting edge and hence increased wear rates. For this reason the run out error has to be minimized in micro milling, because the feed per tooth is in the range of a few microns.

The cutting forces can also give information on the tool deflection, which is much more relevant in micro milling compared to macro milling, as tool stiffness is relatively low. Bending stiffness of less than 100 N/mm can be expected for cutting tools of 100 μm in diameter. Uhlmann and Schauer have shown that tool tip deflection in micro milling can easily exceed 10 μm . For this reason new concepts for the design of micro milling tools were developed, considering the type of dynamic loads micro tools have to withstand. New tool designs are shown in Fig. 7.51. Optimized micro end mills show 30 % less stress encountering the same forces. In fact this reduces tool deflection and also the risk of tool breakage [106].

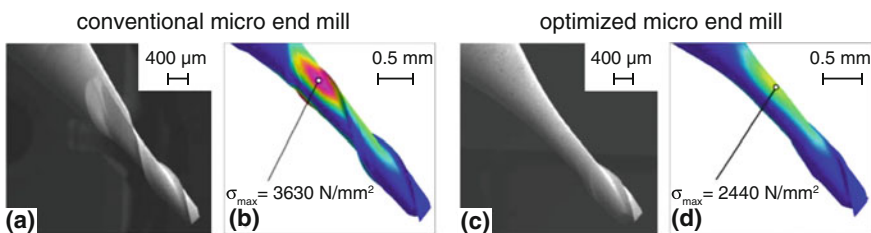


Fig. 7.51 Reduction in maximum stress as consequence of new shank and flute design [106]

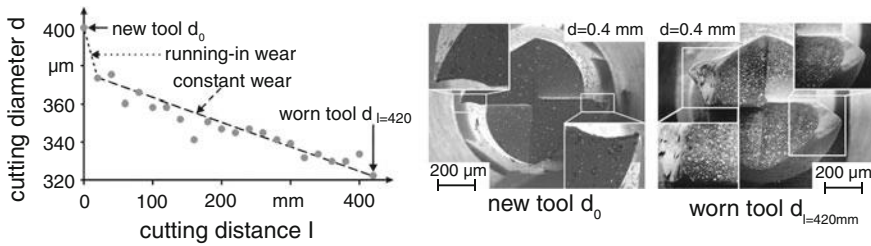


Fig. 7.52 Tool wear of a 400 μm straight end cutting tool in the micro slot milling of hardened tool steel

7.4.2.6 Tool Wear

In micro milling of steel tool, wear is a major problem. Abrasive material removal is the dominating wear mechanism and leads to a continuous reduction in cutting tool diameter during the milling process in the case of slotting or side milling, as shown in Fig. 7.52. It can be recognized that the highest wear rate occurs during the first millimeters of cutting, which can be explained by a very high wear of one cutting edge due to the run out error of the flutes. In addition to form deviations, worn out cutting edges also increase burr formation. For high form accuracy and surface quality it is essential to apply a fresh cutting tool for finishing after roughing.

In ball end milling, if the cutting tool is not inclined, the highest wear rate occurs at the tool tip, where the cutting speed is zero. Figure 7.53 shows a heavily worn micro ball end tool after finishing an area of 12 mm^2 hardened tool steel. The first 16 μm of the tool tip were completely removed by abrasive wear. Machining experiments have shown that tool wear in micro ball end milling cannot be correlated to the workpiece material hardness. It is assumed that the alloying elements involved and the type of carbides are much more influential on the wear rate than the hardness. For this reason it is helpful to compare different workpiece materials—with similar mechanical properties, but different alloy composition—for cost efficient mold making by micro milling, especially if batch sizes increase.

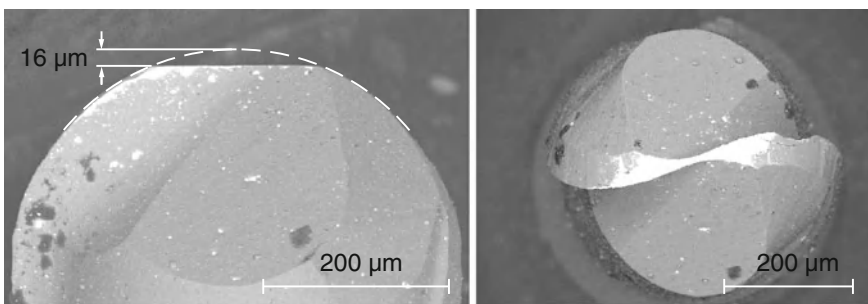


Fig. 7.53 Side and top view of a heavily worn 500 μm micro ball end mill

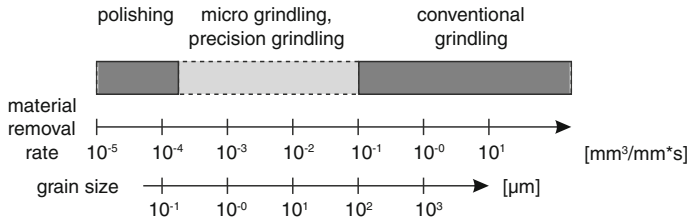


Fig. 7.54 Typical specific material removal rates and applied grain sizes in conventional grinding, precision grinding, micro grinding and polishing (according to [66])

7.4.3 Grinding

Grinding is a machining technology for the shaping and finishing of hard and brittle materials, e.g. hardened steel, cemented carbides or ceramics, which cannot be machined by cutting. In contrast with cutting, grinding is characterized by an undetermined number and orientation of many cutting edges exhibiting negative rake angles, which lead to different material removal mechanisms. In grinding the material removal rate is mostly specified with regard to the grinding tool width, and for this reason called the specific removal rate Q'_w . Figure 7.54 shows typical specific material removal rates and applied grain sizes in micro grinding. It can be recognized that Q'_w is relatively low, and for this reason micro or precision grinding is only suitable as a finish machining technology.

7.4.3.1 Material Removal Mechanisms

Usually the grinding of hard and brittle materials causes microcracks which deteriorate the surface quality. Thus the transition from brittle-to-ductile material removal is considered to be of great importance for micro and precision grinding. Much research effort has been spent to identify this transition and to understand the removal mechanisms [4]. A major parameter for the transition from brittle-to-ductile behaviour in chip removal is the stress condition in the workpiece material around the cutting edge.

From the theory of plasticity and fracture mechanics it is known that the degree of plastic deformation depends on the temperature, strain rate, as well as on the stress state, e.g. multi axial compression or tensile stress in the workpiece. Figure 7.55 illustrates the Coulomb-Mohr hypothesis which describes multi axial compression and tensile stress conditions within crystalline and amorphous workpieces which cannot comply with high tensile stresses [67]. From this hypothesis it can be determined that hydrostatic compression stress fields in the shear plane are necessary for the ductile cutting of hard and brittle materials. Karman has shown that hard and brittle materials can be machined in ductile mode (i.e. visco-plastic material flow) if the hydrostatic compression and shear stresses

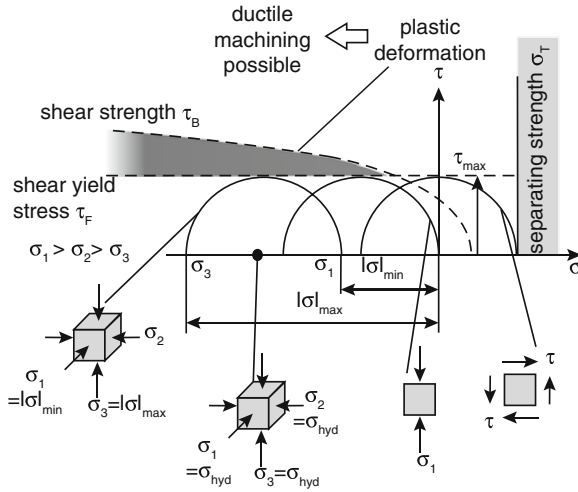


Fig. 7.55 Influence of stress state on shear strength τ_B , shear yield stress τ_F and plastic deformation of hard and brittle materials (according to [67])

are sufficiently high [47]. The contact of the abrasives with the machined substrate leads to elastic material response and, with increasing depth of cut, to plastic behavior, i.e. “micro grooving” and “micro ploughing” (Fig. 7.56a, b). An increasing depth of cut leads to “micro cutting” (Fig. 7.56c). Eventually, a further increase in depth of cut leads to micro-cracks below the workpiece surface (Fig. 7.56d) [90].

Generating hydrostatic pressure and maintaining a critical chip thickness are prerequisites for the crack-free, ductile grinding of brittle materials. Ductile and damage-free grinding of hard and brittle materials requires a maximum chip thickness $h_{cu,max}$ not exceeding the critical and material specific chip thickness $h_{cu,crit}$ to avoid crack initiation [53]. The critical material specific chip thickness $h_{cu,crit}$ depends on the Young’s modulus E , Knoop hardness HK and the critical fracture toughness K_c [60]:

$$h_{cu,crit} \sim \left(\frac{E}{HK} \right) \cdot \left(\frac{K_c}{HK} \right)^2 \quad (7.20)$$

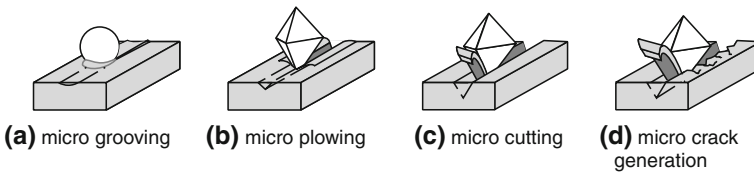


Fig. 7.56 Material removal mechanism in grinding of brittle materials (according to [90])

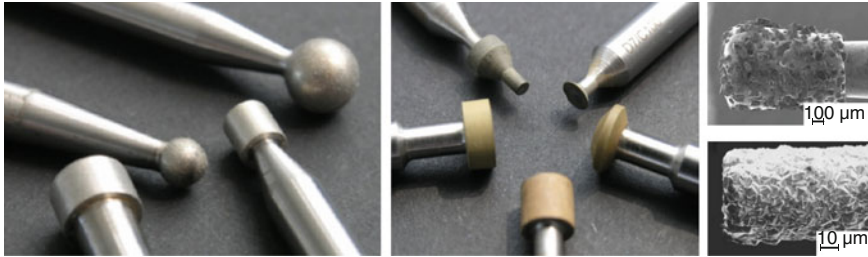


Fig. 7.57 Photographs of diamond grinding tools: electroplated (*left*), resin-bonded (*centre*) and SEM-pictures of diamond micro pins: electroplated (*top right*) and CVD-coated (*bottom right* [29])

7.4.3.2 Grinding Tools

In conventional and also in precision grinding usually tools in the form of grinding wheels are applied. In micro grinding tools of pin type or ball end shape are also common. Micro grinding tools are usually manufactured in the form of a solid body made of steel or cemented carbide with an abrasive layer consisting of a bond and embedded grains, e.g. diamond or CBN grains. Grain sizes in micro grinding are typically smaller than 100 μm . Besides conventional electroplating of diamond grains to the tool body, an alternative technology is emerging, where tool bodies are coated by chemical vapor deposition (CVD) with a rough diamond layer (cf. Fig. 7.57). CVD coated grinding tools have already been manufactured down to a diameter of some ten micro meters. Truing and dressing is challenging for both types of grinding tool, because there is only a mono layer of abrasives.

7.4.3.3 Grinding Kinematics for Mold Making

Besides conventional 5-axes machining, several grinding kinematics especially for the manufacturing of rotational symmetric molds were developed in recent years. According to Tohme [101] cross axis grinding (see Fig. 7.58a) is the most common 2-axes grinding technique, which is useful for the grinding of molds and dies. Here, the wheel rotational direction is transverse to the rotational direction of the ground workpiece. The grinding of the part is carried out by a simultaneous 2-axis movement. The disadvantage of the cross axis grinding technique, especially in mold manufacturing, is the size of the grinding wheel and the required clearance. In order to avoid collisions, tool radius must be greater than workpiece's surface maximum sag when machining concave molds. Also, the shaft length must be larger than the workpiece radius to provide full access; unfortunately, long tool shafts in combination with high process forces tend towards high deflections resulting in dynamic problems and high figure errors of the ground workpiece. To overcome the disadvantage of the cross grinding method, tilted or inclined

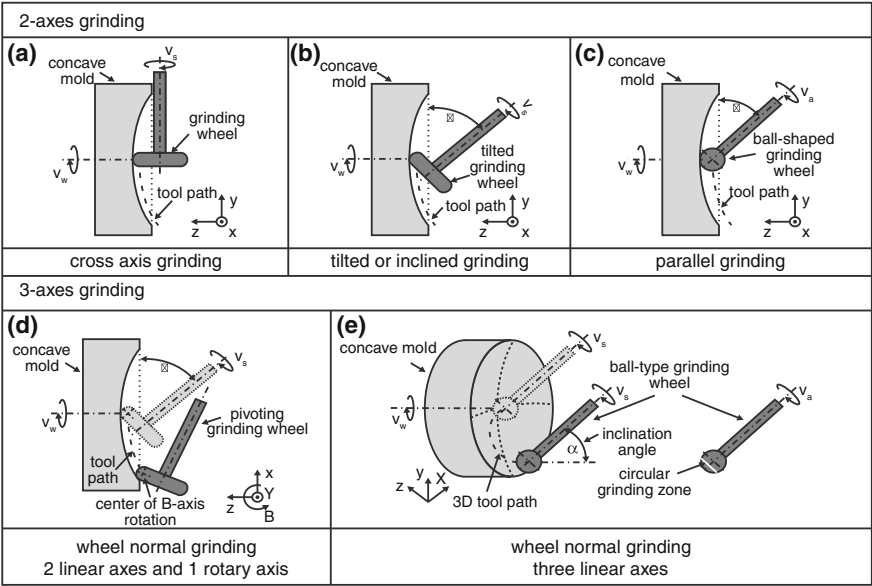


Fig. 7.58 Overview of different grinding kinematics for mold manufacturing sorted by the number of axes used [10]

grinding (see Fig. 7.58b) has been developed. Again, the feed movement is performed in two axes. Distinct from cross axis grinding, the grinding spindle rotational axis is inclined with an angle α of e.g. 45° relative to the workpiece surface. This enables flexible access to the workpiece surface and tool designs with smaller shaft lengths, increasing the setup stiffness and the form accuracy achievable.

In order to increase the grinding tool life and to reduce the complexity of tool compensation methods, the parallel grinding (see Fig. 7.58c) process was developed and analyzed by Saeki et al. [83]. Here, a spherical or ball-shaped grinding wheel is tilted horizontally by 45° with respect to the workpiece's rotational axis and moves across the part using two simultaneous axes of motion. During the grinding process the rotational direction of the workpiece and the cutting direction of the wheel become parallel at the grinding point which moves along the grinding wheel profile. Thus wheel wear can be reduced and the tool life is improved. A disadvantage of this grinding technique is the sensitivity towards wheel geometry errors because the grinding wheel shape is directly transferred to the ground surface and affects the form accuracy of the mold. For this reason, once again precision wheel truing and dressing processes as well as continuous workpiece form measurement with a robust form error compensation technique are required for the achievement of the required high surface form accuracy.

An improvement of the parallel grinding technique was developed by means of the wheel normal grinding method (see Fig. 7.58d and e). Here, the grinding wheel contact point is kept constant across the entire workpiece surface, which ensures

that the geometric tool errors will not be copied into the part during the grinding process. This type of grinding process requires three axes of motion. It also enables the use of a variety of different tool geometries, like for e.g. spherical micro grinding wheels for aspherical shaped, concave mold cavities or sharp edged cylindrical wheels for the generation of small blend radii between convex and flat functional surfaces or micro structured surfaces.

Tohme [101] reported on a wheel normal grinding process using two linear axes and a rotary axis. In this case the grinding wheel is kept normal to the workpiece surface by commanding the rotary B-axis to remain perpendicular to the point of contact. Therefore, the included angle α between grinding spindle axis and workpiece surface needs to be controlled over the entire grinding process (see Fig. 7.58d). Under these requirements the figure accuracy of the ground surface is minimized to selective tool wear because wheel form errors do not copy onto the ground mold.

7.4.3.4 Surface Quality

Surface quality in terms of roughness results from the machining parameters as well as from the grain size of the applied abrasives. For a smooth surface finish the grain size has to be small combined with a soft bond. In micro grinding surface qualities of $S_a < 100$ nm can be achieved under the application of ultra precision machine tools and very fine grained grinding tools, even $S_a < 10$ nm. For high form accuracy the bond has to be harder to reduce form deviations resulting from tool deformation.

7.4.3.5 Ultrasonic Assisted Grinding

Wheel speed is one of the most influential parameters in grinding. If the grinding process is scaled down, the wheel speed decreases proportional to the tool diameter. For this reason very high spindle speeds are required in micro grinding, especially if the cutting tool diameter is in the range of a few millimeters or even less. Besides high wheel speeds, another approach to increase the material removal rate is the superposition of ultrasonic vibration. In ultrasonic assisted grinding an additional axial-movement with low amplitude (0.5–2 μm) but high frequency (20–40 kHz) is superimposed onto the rotational velocity. This motion has two effects, on the one hand the effective cutting speed increases and on the other the axial movement operates as a micro jackhammer, shattering the surface layer. Figure 7.59 shows the result of grinding experiments with and without ultrasonic assistance. It can be seen that the achievable surface quality in ultrasonic assisted grinding is almost independent of the removal rate. If no ultrasound is applied, the surface quality decreases with increasing removal rate. Therefore, ultrasonic assisted grinding is useful for producing roughing with very high removal rates,

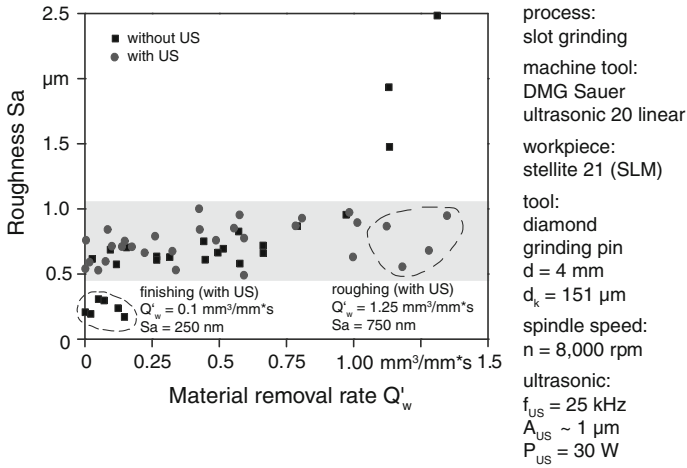


Fig. 7.59 Comparing material removal rate versus surface quality in grinding of satellite 21 with and without ultrasonic assistance

but at low removal rates and without ultrasonic assistance a better surface finish can be achieved.

7.4.4 Polishing

The classification of manufacturing processes according to DIN 8580 does not include polishing as a specified manufacturing process. Due to the complexity of the process and the numerous mechanisms of action, a precise definition of polishing is difficult. In some publications polishing is described as a process for generating a minimal surfaces roughness with high form accuracy and a surface layer with no subsurface damage. Within the process chain for the manufacturing of optical or technical surfaces, polishing is usually the last machining step for reaching a very low surface roughness, typically in the nanometer range. In mold making for micro forming processes polishing is an important finishing step, especially in micro forming processes in which the surface topography of the forming tools is directly transferred onto the formed part, like in micro rotary swaging.

7.4.4.1 Abrasive Polishing

In mechanical abrasive polishing processes the material is removed through an abrasive action of grains which are supplied to the process in a polishing suspension. Typical abrasives are mono- or polycrystalline diamond, because of their high hardness and wear stability. In abrasive polishing a tool in the common sense

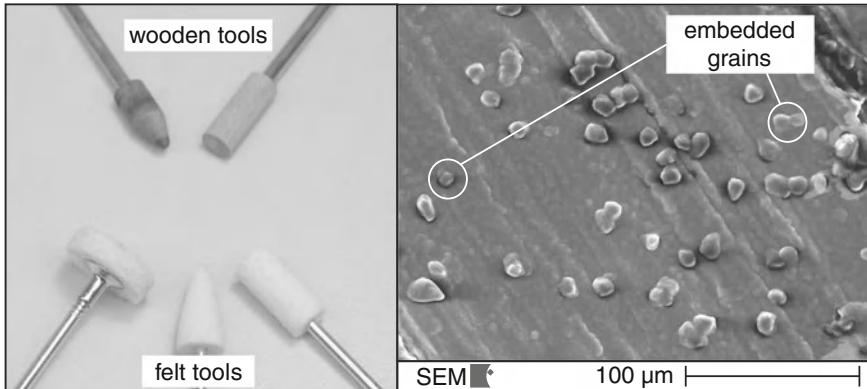


Fig. 7.60 Different types of polishing tool and a SEM-image of a polishing tool surface with diamond grains partly embedded in the pad material (wood)

does not exist, as the tool is emerging in the process from the interaction of the grains, i.e. the suspension, the pad carrier and the workpiece.

In Fig. 7.60 different rotational polishing tools are shown. Depending on the hardness of the pad material the removal rate varies due to the different embedding of the grains in the pad. On the right side an SEM-image shows diamond grains with a grain size of $9\ \mu\text{m}$ on a wooden polishing pad after the polishing of hardened steel. Grains are partly embedded in the pad material as well as lying on the pad surface.

Relative motion between the pad carrier and the workpiece leads to material removal. In abrasive polishing two abrasive mechanisms govern the process, shown in Fig. 7.61. On the one hand there is two-body abrasion between the active grains and the workpiece. This removal mechanism is characterized by embedded grains in the polishing pad which are forced over the surface in a constrained motion. In this case a microcutting or microploughing removal mechanism is dominant. The surface of the workpiece typically shows a structure like grinding with directional cutting grooves, as shown in Fig. 7.61 left. On the other hand grains can roll over the workpiece. This so-called three-body abrasion is characterized by the generation of microcracks or microfatigue resulting in the removal of material by small particles from the workpiece's surface. In this case the surface exhibits a typically undirected structure with statistically distributed pits of removed material, as shown in Fig. 7.61 right. The dominating removal mechanism mainly depends on the contact force of the polishing tool and workpiece and the properties of the pad material. The embedding of abrasive grains occurs more easily in soft pad materials, and the two-body abrasion is dominant. However, a higher force facilitates the embedding of grains in the pad. With hard pad materials and low forces a rolling of abrasives over the surface is more likely and material removal by microfatigue dominates.

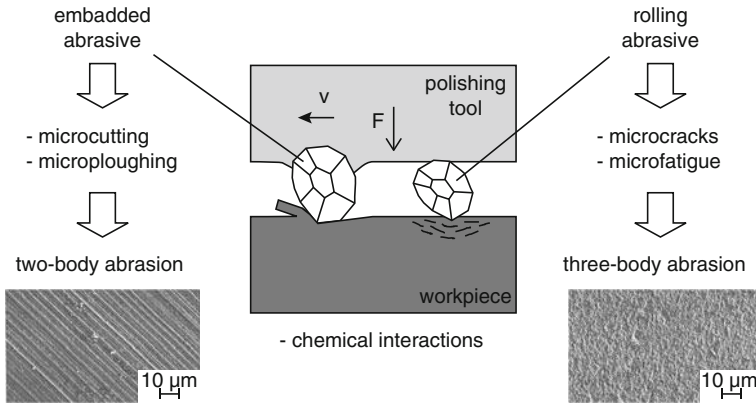


Fig. 7.61 Material removal mechanisms in abrasive polishing processes [8]

The characterization of an abrasive polishing process was already investigated by Preston in 1927 [83]. Preston developed a hypothesis for the temporal dependence of material removal from three factors:

- polishing pressure p
- relative velocity between workpiece and polishing tool v_{rel} and
- a constant factor K_p , the Preston constant.

$$\frac{dz}{dt} = K_p \cdot p \cdot v_{\text{rel}} \quad (7.21)$$

Except for pressure p , polishing time t and relative velocity v_{rel} , the Preston constant summarizes all remaining factors influencing material removal:

- polishing tool
- polishing abrasive
- polishing suspension
- machine tool
- workpiece and
- process behavior.

For the finish machining of small molds or dies polishing with adapted pad shapes is essential, the so-called profile polishing. Depending on the structure to be polished, different kinematics can be applied. Figure 7.62 shows an example of a profile polishing process for finish machining of the forming zone of a micro rotary swaging tool. Micro rotary swaging is a forging process in which the surface topography of the forming tool is directly copied onto the formed part. For this reason a high quality surface finish is essential for high part quality. In this example the rotary swaging tool was premachined by micro ball end milling, achieving a surface roughness of $S_a = 150\ \text{nm}$ and afterwards polished. The

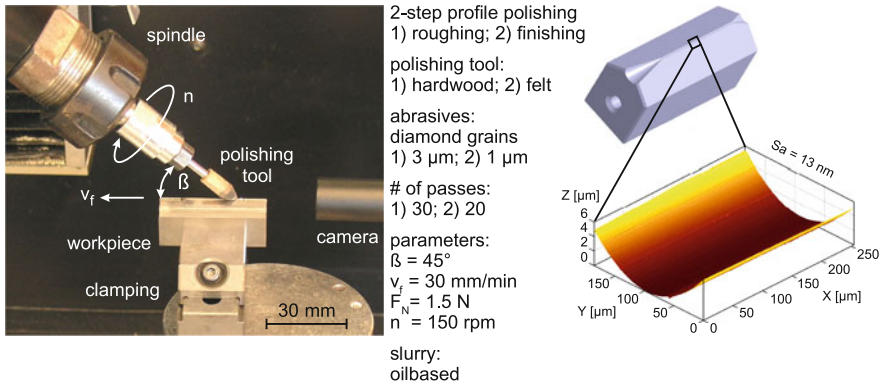


Fig. 7.62 Profile polishing of a micro rotary swaging tool

polishing process was performed in 2 steps, roughing with a hardwood polishing tool and 3 μm diamond grains and finishing with a felt tool and 1 μm diamond grains. After polishing a surface roughness of $S_a = 13 \text{ nm}$ was achieved in the forming zone. The process setup and applied parameters are shown in Fig. 7.62.

7.4.4.2 Laser Polishing

Laser polishing is comparable with a multistage system of premachining and finishing. The laser beam treats the workpiece surface under a certain angle and melts a thin surface layer depending on the adjusted power of the laser, shown in Fig. 7.63 (left). The following solidification of the material results in a leveling of the roughness peaks, due to an aspiring of minimal surface energy of the remolten surface layer [70]. In the first machining step, macro polishing, a leveling of deep structures from a previous machining step e.g. milling or EDM-processes is achieved by continuous laser radiation. The surface layer, which is molten, ranges from 40 to 80 μm . During the second machining step, micro polishing, the micro roughness remelts and micro roughness peaks evaporate by applying a pulsed laser radiation. In this case the thickness of the molten layer is about 5 μm , so the surface quality and especially the glossiness will be further improved. In addition to the angle of incidence, the power and the mode of the laser (pulsed or continuous), beam diameter, track spacing and feed rate are the most important process parameters. Typically the laser rasters in a meandering pattern over the surface [70]. Figure 7.63 (right) shows a white light interferometer measurement of a laser polished steel surface. Roughness R_a of the laser polished structures is in the range of 400–600 nm [7].

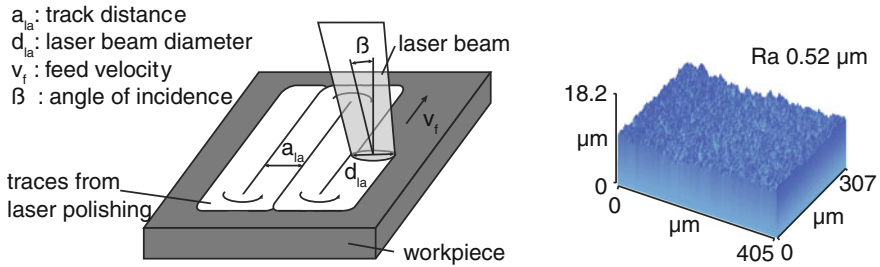


Fig. 7.63 Principle of laser polishing after [70] (left) and white light interferometer measurement of a laser polished surface (right) [7]

7.4.4.3 Friction Polishing

In the previous chapter, diamond is used as an abrasive material for polishing because of its exceeding properties such as high hardness, low friction, high chemical inertness and high wear resistance. Because of these properties diamond appears to be an interesting candidate for micro forming tools to shape advanced materials like e.g. aluminum-scandium-alloys by metal forming [6]. In contrast with natural diamonds, synthetic diamonds are monocrystalline or polycrystalline and almost free of inclusions. Therefore, the objective of the technology introduced here, called friction polishing, aims at the final machining of pre-machined synthetic diamond workpieces as mold inserts.

The basic principle of friction polishing is based on a thermochemical process in-between the diamond workpiece and a friction rod. To initialize a thermochemical reaction between the diamond's carbon, friction rod and ambient oxygen, high contact temperatures are required. Frictional heat is generated in the contact area between the rod and workpiece, resulting from relative motion and applied force. In any case, due to excellent thermal conductivity and the associated heat flux from the effective area, additional heating of the workpiece is required. During the machining a continuous removal on the rod's surface is observed and nearly carbon-free material is working at the contact point. Given that the friction rod is a tool for micro machining, the revolution of the rod has to be high in order to achieve an effective relative velocity. Figure 7.64 shows the principle design of the friction polishing process with typical parameters and a polished diamond surface.

Up until now, the reaction-based wear of diamonds in contact with steel is not fully understood. Paul et al. investigated the effects of diamond wear on different metals. Whereby the authors described, "that the relevant factor in chemical wear is the presence of unpaired "d" electrons in the sample" [74]. These atoms crack the close bonds between the carbon-carbon atoms inside the diamond lattice. After losing the close bonds the free carbon atoms are in a higher reactive phase. In the second step the authors propose four different reaction mechanisms of these carbon atoms:

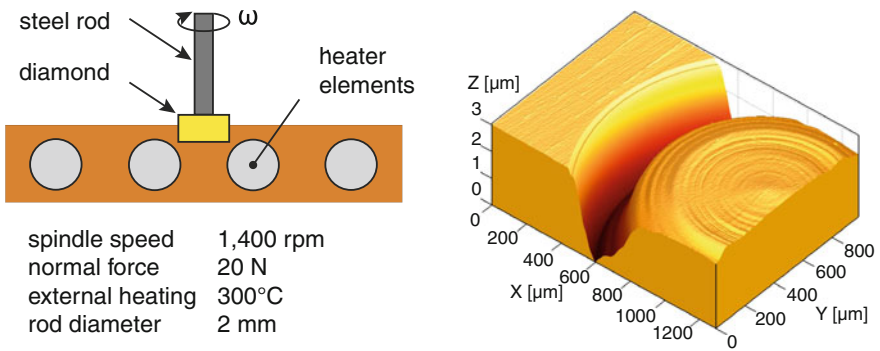


Fig. 7.64 Setup of friction polishing (*left*) and polished diamond surface (*right*)

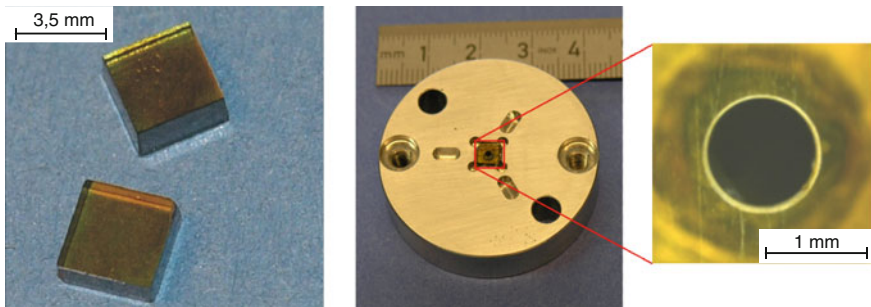


Fig. 7.65 Synthetic monocrystalline diamonds (*left*) and machined diamond as a forming die (*right*)

- diffusion into the counterpart (here: friction rod)
- formation of graphite on the diamond (here: mold insert)
- reaction with oxygen to form CO or CO₂
- reaction with the workpiece (here: mold insert) to form carbides.

These reactions are utilized for the shaping of diamond in friction polishing. Figure 7.65 shows two synthetic generated monocrystalline diamonds as well as a diamond forming die inside a tool holder. To provide the basic geometry the diamond was machined by laser cutting and subsequently friction polished inside the cylindrical pilot hole. With this diamond forming die small micro cups with a diameter of 1 mm can be produced without the application of lubricants.

7.5 Surface Coating

M. Schwander and K. Partes

7.5.1 Physical Vapor deposition

Physical vapor deposition (PVD) summarizes coating processes, which deposits thin films by vaporizing solid or liquid materials and then directly deposits them onto surfaces. As shown in Fig. 7.66, a solid can be transformed into vapor by melting and then evaporation or sublimation [62]. An important difference with other coating processes is the direct deposition from the vapor phase without reaction at the surface, or condensation and solidification. The material, which is designated the target, can vaporize in two different ways. First, the target can be heated until the required steam pressure is reached. This can be done by resistance heating or induction heating, for example. Second, by bombardment with highly energetic ions, or electrons or photons by laser, the desired material can be ejected from the target. All materials which do not degrade by the transition from the solid to gaseous state can be deposited by using a PVD process.

An example of a typical PVD reactor is given in Fig. 7.67. The pressure inside the chamber is usually reduced by a multi-stage pumping system in the range of 10^{-3} – 10^{-2} Pa. Typical PVD deposition rates vary from 10 to 100 Å/sec [62]. Since the target can consist of a powder or fluid, it is usually kept at the bottom of the chamber in a container. In the illustrated case, the energy supply is fed by a resistance heater. The vaporization rate depends on the temperature and the equilibrium vapor pressure of the material. For materials with a high vapor pressure at low temperatures, a resistant heater is commonly used for evaporation. However, for high temperature materials, which require a high temperature to vaporize, a focused high-energy electron or laser beam system is often used. Depending on the material and the energy supply, the vaporized material consists

Fig. 7.66 Phase diagram with the designated area of PVD

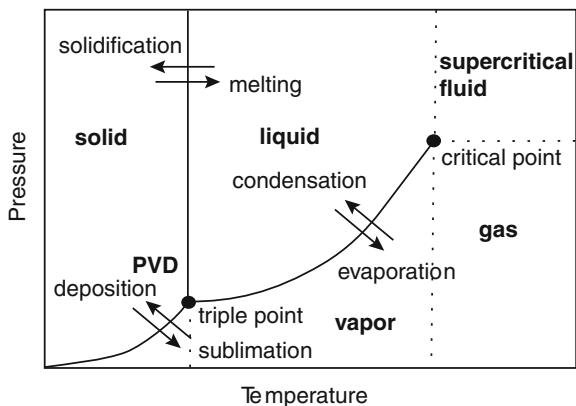
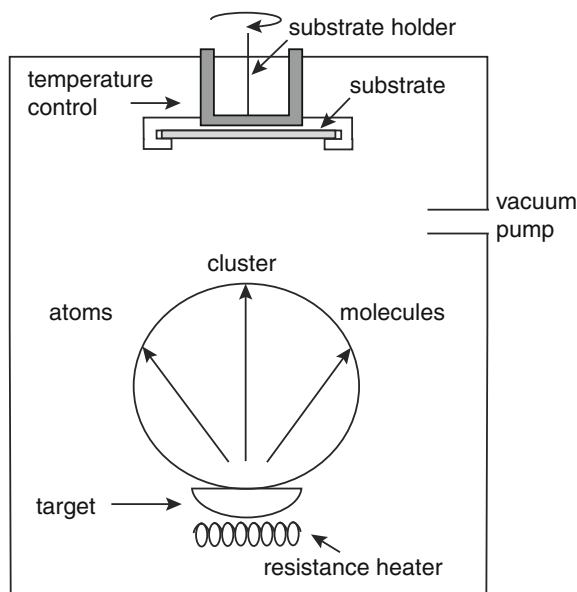


Fig. 7.67 Principle concept of a PVD reactor

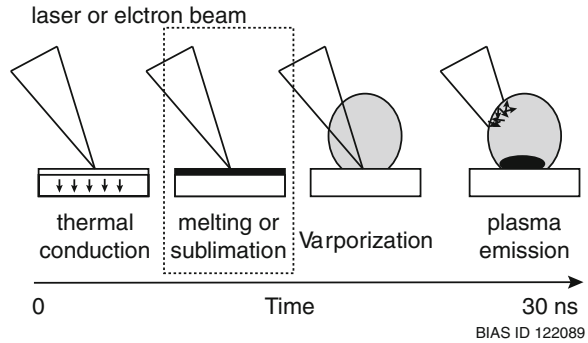


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of atoms, clusters or molecules. Especially antimony, carbon and selenium have the tendency to generate clusters. Therefore, these materials are usually vaporized by a source called a baffle source, which ensures the depositing vapor is in the form of atoms. Besides evaporation as atoms and clusters, some materials primarily vaporize as molecules. The degree of atoms and molecules of a compound depends strongly on the vaporization temperature and the degree of dissociation. Through the low pressure and high temperature of the target, the evaporated material moves through the chamber in the direction of the sample. Depending on the shape of the target, the energy supply and the distance to the sample, evaporation can be seen as a point source or a material beam. In the upper region of the chamber, the sample is fixed face down on a rotating and temperature-controlled disk. As shown in Fig. 7.67, the vaporized material can be deposited through the cooling of the sample.

As already stated, the electron beam and laser beam evaporation are two commonly used methods for the energy supply. The basic operating principle of laser beam evaporation is shown in Fig. 7.68. In the first nanoseconds (ns) the laser beam hits the target. Depending on the material, wavelength and angle of incidence between the beam and surface normal, the laser radiation is absorbed or reflected. The absorbed power can either melt the surface and then evaporate, or directly sublimate from solid into vapor. Another possibility, which is not shown, is that depending on the thermal stresses introduced, individual clusters sputter out of the target. The material which is converted into vapor increases the density in front of the target and can absorb, for example via multiphoton absorption, the laser radiation. By increasing the energy density the vapor becomes ionized and turns into a

Fig. 7.68 Time course of vaporization using a laser or electron beam

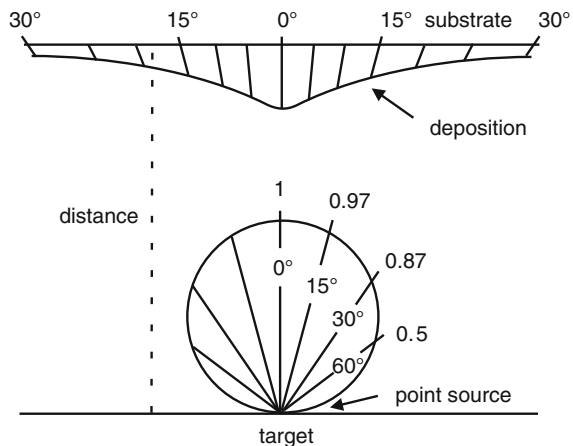


plasma. Since plasma consists of charged particles, more possibilities of interaction develop, such as the inverse bremsstrahlung, which increases the absorption.

A disadvantage of this kind of energy supply is punctuate evaporation on a small area of the target. A point source results in a cosine distribution of the vaporized materials, as shown in Fig. 7.69 [62]. The cosine distribution describes a coating which has a maximum deposition in the direction normal to the surface and a minimum of 0 at an angle parallel to the plane. Without moving the sample, the resulting coating is mainly dependent on the distance and substrate orientation according to the cosine source. In advanced coating systems the flux distribution deviates from the cosine source through modified source geometry, collisions in the vapor above the target, changes in energy source geometry, etc. However, during PVD processes the depositing material is always in line-of-sight of the target.

The deposition of a thin film which uniformly coats a three-dimensional structure is known as conformal film growth. The conformality and homogeneity of a PVD coating depends primarily on the particle flux, the surface structure and the reactive sticking probability. In a small area near the surface, the shape of the evaporation source can be neglected. Therefore the particle flux can be assumed as

Fig. 7.69 Distribution of vaporized and deposited material by a point source. Based on [62]



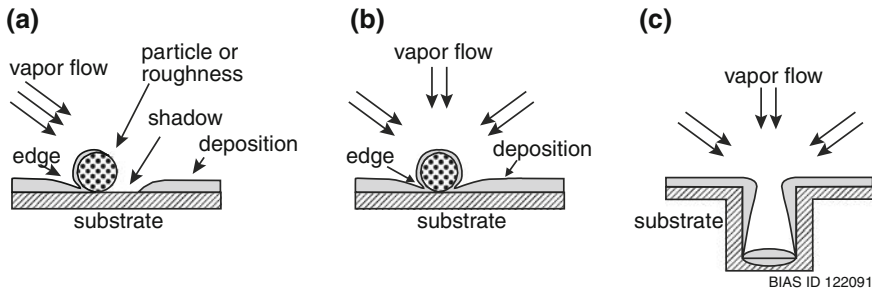


Fig. 7.70 Coating defects that can occur in PVD processes by geometrical effects. Based on [63, Chap. 10]

a coherent beam. Figure 7.70a shows an example of the effect that surface morphology and particulate contamination can have on surface coverage by a simple particle flux. Depending on the angle of the incoming particle flow, roughness peaks and contamination shadows can be generated, where no deposition occurs. To reduce the shadows, the sample can be rotated, whereby the coating can occur equally from all sides. However, there are still areas with different layer thicknesses as shown in Fig. 7.70b. A far greater impact than particle flux and surface structure on conformal growth is the reactive sticking probability S_R . It represents the ratio between the incident molecules (atoms) and the number of sticking molecules on a specified surface. A maximum value of 1 is achieved when every collision results in deposition. The non-conformal growth shown in Fig. 7.70c is obtained using a typical PVD process with a line-of-sight source and $S_R \sim 1$. Due to preferential growth near the top of the side walls and on the bottom surface, the flux towards the side walls near the bottom becomes reduced. This result is confirmed by calculated profiles and can be found by evaporation or sputtering processes. These issues mean PVD processes become very critical for integrated circuits with feature sizes less than $0.5\ \mu\text{m}$, because structures with rather high aspect ratios must be coated uniformly or filled.

Beside these geometric effects, coated surfaces contain grooves, ridges, conical particles (hillocks), pin-holes, pores and open voids. The most identified mechanism for the generation of surface defects is a contaminated surface. In particular, particles which are stuck to the surface before the coating process can lead to flake defects with typical dimension of $5\text{--}40\ \mu\text{m}$ [72]. Sources of these defects can be mechanical pre-treatment or etching processes [13]. Furthermore, fine particles can originate during the deposition process from vacuum chamber components as well through the formation of dust particles in plasma. These defects are, with an average size of $1\ \mu\text{m}$, substantially smaller than flake defects. Since defects are drawbacks in the area of micro forming, it is important to minimize the defect density. However, the number of possible reasons for defects occurring is very large. In general it is nearly impossible to generate coatings without any defects [73]. Therefore the reduction in defect density usually only occurs through parameter studies on selected coating processes [1].

7.5.2 Chemical Vapor Deposition

In contrast to physical vapor deposition (PVD), chemical vapor deposition (CVD) cannot be described by a phase transformation from vapor to solid. CVD processes involve a chemical reaction of the gas above the surface before the precursor is deposited as powder or film on the substrate surface. The gas used as a starting material for the chemical reaction will be referred to as a precursor. Typical precursors are metals, metal hydrides, halides, halo hydrides and metal organic compounds.

The industrial use of CVD processes starts with the patent by de Lodguine in 1893 for the deposition of W onto carbon lamp filaments through the reduction of WCl_6 by H_2 [21]. The principle coating process and the key steps in the majority CVD processes are illustrated in Fig. 7.71. The individual points represent:

1. Generation of the precursors out of the solid or liquid phase through reaction with a gas or evaporation. In most cases this point is omitted, since the precursors are delivered in gaseous form.
2. Diffusion and transport of the precursors inside a reaction chamber, or a nozzle.
3. Through thermal heating or ionization the precursor can undergo subsequent decomposition or chemical reaction.
4. Through a reaction at the interface between gas and solid the precursor becomes absorbed and produces a deposit. The chemical reaction may also include the growth and etching of several intermediates. Requirements for deposition include: temperature of the surface (i.e. heating), temperature of the precursors, excitation level and decomposition of the precursors. A variety of chemical reactions are given in Table 7.7.
5. After absorption the deposited molecules can diffuse along the surface, decomposed and desorbed, or surface reactions can take place, and a film can form with various side products.
6. By-products are removed from the chamber by diffusion or convection.

Fig. 7.71 Schematic illustration of the operating process during CVD deposition

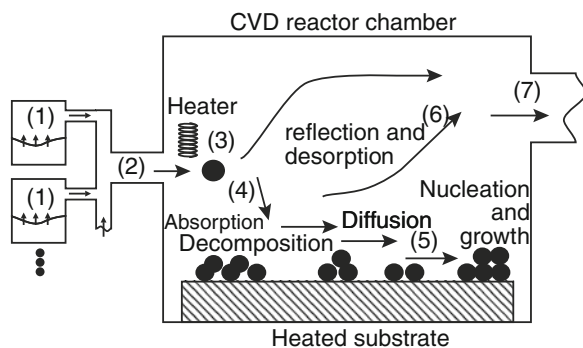


Table 7.7 Examples of chemical processes used for CVD reactions

CVD reaction	Chemical precursor	Example	T _{dep} [°C]
Thermal decomposition	• Halides	$\text{TiI}_4 \rightarrow \text{Ti} + 2\text{I}_2$	1,200
	• Hydrides	$\text{SiH}_4 \rightarrow \text{Si} + 2\text{H}_2$	600–1,150
	• Metal carbonyl		
	• Metalorganic		
Reduction	• Halides	$\text{SiCl}_4 \rightleftharpoons \text{Si} + 4\text{HCl}$	900–1,200
		$\text{WF}_6 \rightleftharpoons \text{W} + 6\text{HF}$	550–800
Oxidation	• Halides	$\text{SiH}_4 + 2\text{O}_2 \rightarrow \text{SiO}_2 + 2\text{H}_2\text{O}$	350
	• Hydrides	$\text{Zn}(\text{C}_2\text{H}_5)_2 + 4\text{O}_2 \rightarrow \text{ZnO} + 5\text{H}_2\text{O}$	250–500
	• Metalorganic		
Hydrolysis	• Halides	$2\text{AlCl}_3 + 3\text{CO}_2 + 3\text{H}_2 \rightarrow$	800–1,150
		$\text{Al}_2\text{O}_3 + 6\text{HCl} + 3\text{CO}$	
Nitridation	• Halides	$\text{TiCl}_4 + 1/2\text{N}_2 + 2\text{H}_2 \rightarrow \text{TiN} + 4\text{HCl}$	1,200
	• Hydrides	$3\text{SiH}_4 + 2\text{N}_2\text{H}_4 \rightarrow \text{Si}_3\text{N}_4 + 10\text{H}_2$	600–1,100
	• Halohydrides		
Disproportionation	• Halides	$\text{GeI}_2 \rightleftharpoons \text{Ge} + \text{GeI}_4$	
		$\text{Si} + \text{SiI}_4 \rightleftharpoons 2\text{SiI}_2$	
Synthesis	Two or more precursors	$\text{TiCl}_4 + 2\text{BCl}_3 + 5\text{H}_2 \rightarrow \text{TiB}_2 + 10\text{HCl}$	1,100
		$\text{CH}_4 + \text{H}_2 \rightarrow \text{C} + 3\text{H}_2$	900–1,100

7. Depending on the toxicity and environmental impact of the by-products and unreacted precursors the gas will be transported out into the atmosphere or will be recycled.

A very simple example of a CVD process is the “Mond process”. The procedure, which is named after the chemist Ludwig Mond, is mainly used for the cleaning of nickel but can also be used to deposit coatings on a heated substrate. The heating of nickel to 50–60 °C inside a carbon monoxide (CO) atmosphere produces tetracarbonylnickel as the precursor. Impurities like cobalt and iron do not form carbonyls. Hence these contaminations will turn into the precursor gas. They will remain at the starting point. Without adding additional components, the precursor is transported inside a chamber with a 220–250 °C heated sample. The precursor decomposes through the higher temperatures at the surface back into nickel and carbon monoxide. Nickel is deposited on the substrate and carbon monoxide can be carried away [14]. The purity of the deposition is higher in comparison to the reactant.

In addition to this simple procedure, there are countless different precursors in different CVD-processes. However, the main CVD processes can be separated into the reactions: thermal decomposition, reduction, oxidation, hydrolysis, nitridation, disproportionation and synthesis. An overview of this reaction with a selected example is given in Table 7.7.

As shown in points 4 and 5 of Fig. 7.71, deposition by CVD-processes is dependent on the chemical absorption at the surface. In contrast to the PVD process, the gas flow is not considered as a particle beam. Therefore, the surface

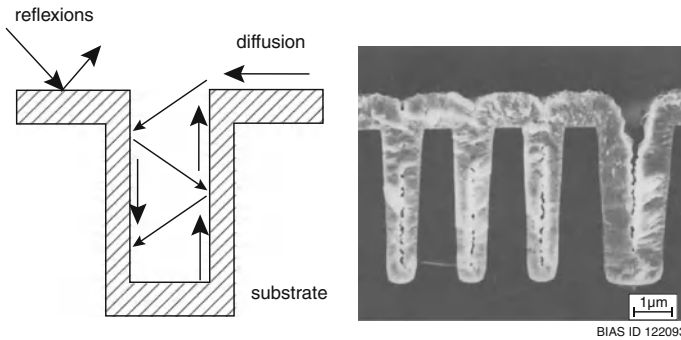


Fig. 7.72 Typical conformal growth of CVD coatings. Based on [24] and [55]

morphology does not have such a strong influence on the layer formation. The high temperature, the high density and the chemical diffusion processes support conformal film growth. In addition, the reactive sticking probability S_R for CVD processes is comparably small; around 0.05. As shown in Fig. 7.72, the control of the chemical reaction at the surface during CVD allows conformal film growth. However, if the precursor was highly reactive, CVD processes can also have a large S_R and hence a high directionality, as shown in Fig. 7.70a). In this case S_R has to be controlled by precursor design or temperature distribution. For example, inert gases such as nitrogen and argon are often used as carrier gases to facilitate the movement of the precursor toward the reaction chamber. By the increase in the degree of carrier gas the reactive sticking probability can be reduced.

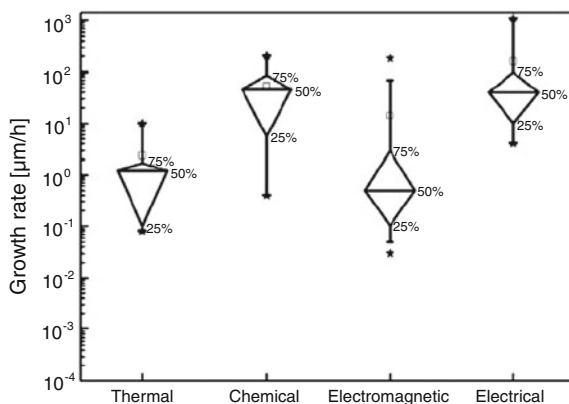
The good aspect ratio presented by CVD coatings are in contrast to PVD coatings mean they are particularly suitable for rough surface structures. However, it has to be considered that the growth of CVD films usually starts with seed crystals on the surface. The distance of the crystal seeds depends on the pre-treatment, the growth material and the substrate. Under some circumstances the distance can be several microns. Therefore, the deposition of CVD films with only a few nm thicknesses is sometimes technically complicated.

Although the growth of CVD films depends on the chemical process, the growth rate depends more on the kind of process and the process parameters. In Fig. 7.73 it can be seen that the growth rate for polycrystalline CVD diamond films can change by up to five orders of magnitude by only changing the energy source and the total gas flow [89].

7.5.3 Multi-Layer Systems

The first coatings have been used since the beginning of the last century in the field of optical coatings. However, the first series production of coatings was used in electronics as a resistor. The enormous potential in electrical engineering resulted

Fig. 7.73 Statistical view of growth rate for different CVD processes. Maximum and minimum values are given by the stars; whiskers gone from 5 to 95 % probability; diamond forms show 25, 50 and 75 % probabilities; the square represents the arithmetic mean [89]



in the rapid development of various coating techniques and materials. Since 1970, coatings on cutting tools were used with the aim of reducing wear and thereby to increasing the lifetime. For several years only local hardening, nitriding and hard chromium processes were used for dies and punches in forming applications. However, the requirements from new materials with high strengths in the field of cutting tools have generally improved the development and application of new advanced PVD and CVD coatings. In contrast with this, the majority of forming tools are still uncoated. This is due to the large size and complex shape of most forming tools. Furthermore, there always remains a residual risk that the coatings peel off from the substrate, even if the coatings obtain a good adhesion. In the case of cuttings tools this can produce difficulties, however not usually accompanied by a disaster. If the coating of a forming tool fails, the detaching material can lead to impairment in product surface quality and a destruction of a very expensive tool. Repairing forming tools by re-coating is usually a very expensive procedure. As for cutting tools, tools for micro metal forming can be also easily repaired by re-coating, since they have the same dimensional order of magnitude. However, this also imposes that the required accuracy of the layers, and the coating process influences on the workpiece must satisfy the surface quality and geometrical accuracy. In the case of deep-drawing, the coated punch should not influence the clearance size, which is only a few micrometers. Hence the use of epitaxial growth crystalline layers is often complicated due to the high temperatures, around 1,000 °C, involved in the deposition process which can lead to the distortion of the component. Additionally, those layers often have a high thickness and high surface roughness which might lead to damage of the cavity.

The first coatings, which were developed as wear protection, have been chosen due to their high hardness. It was assumed that a high hardness goes hand in hand with a high wear resistance. The use of hard coatings resulted in an improved tool life, higher processing speeds and a reduction in lubricant usage. To reduce the use of lubricant, the coatings must have good frictional properties in addition to high wear resistance. Higher processing speeds and fewer lubricants usually leads to higher temperatures, which provides further challenges to the coatings. Current

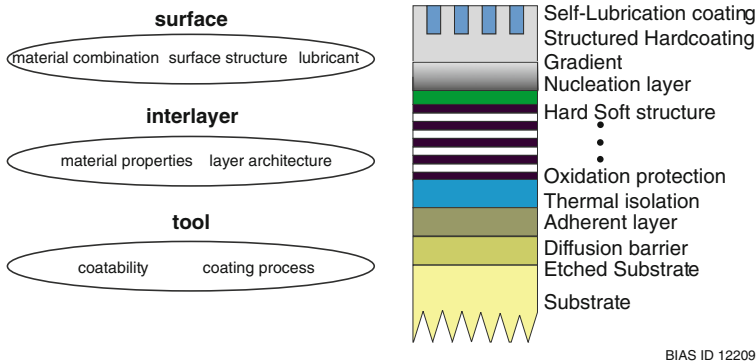


Fig. 7.74 Research areas for multilayers and the application of different coatings in the range of micro metal forming

developments are moving away from single-layer hard coatings, in the direction of high-temperature and oxidation resistant complex layer architectures. The improved mechanical, thermal and physical properties of the coatings move the weak points of the tools away from the surface in the direction of the layers and the base material. Based on this, most research activities can be divided into three areas of interest: surface, interlayer and tool, as shown in Fig. 7.74. Surface includes layers which come in direct contact with the workpiece. The main features are the surface structure, the pairing materials and the lubricants. The interlayer is the connection between the workpiece and hard coating. In most cases they are indispensable in producing a coating or to bond the top coating to the base metal. The main features are the material properties and the architecture of the layers. The field of tools is dominated by the coating process, the tool materials and the resultant coatability.

7.5.3.1 Material Combination

As shown in Sect. 2.2.2, the tool surface wears out due to mechanical and chemical stress in the tribological system of two body interaction. Therefore, the first logical approach is to deposit a layer on the tool which adjusts the properties of the surface of the respective strain. The first coatings that were used were single-layer coatings. At the start they were primarily designed to be hard in order to protect the tool against wear. Therefore, the greatest number of coatings in metal forming are in the hard material coating class. A selection of hard materials, their chemical bonds, and their hardnesses are shown in Figs. 7.75 and 7.76.

If the only factor to consider is the hardness of the material, nothing other than carbon or boron nitride layers will be used, because these alone determine the area of greatest hardness. Meanwhile the widely used TiN layer does not stand out with a hardness of 2,500 HV0.05. In addition to hard coatings, the hardness of the

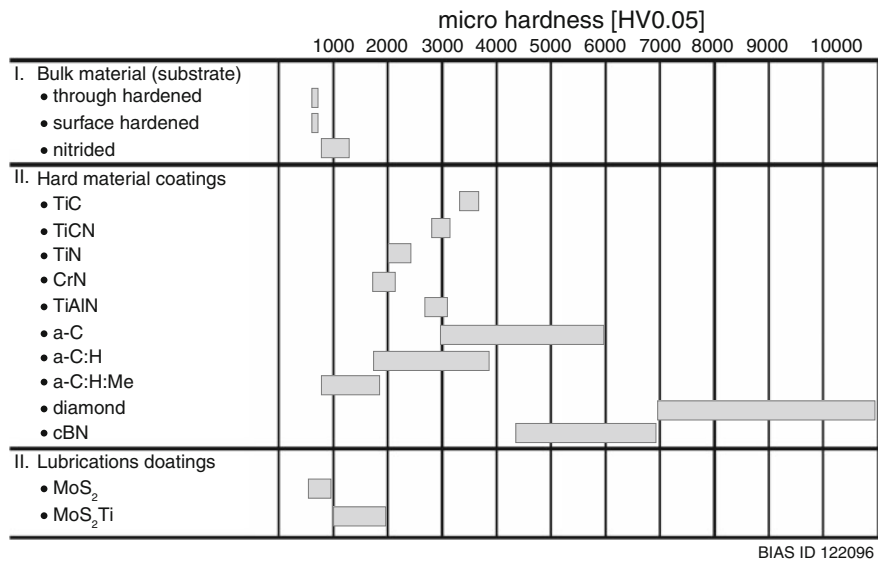


Fig. 7.75 Hardness of coatings and surface treatments used as final layers

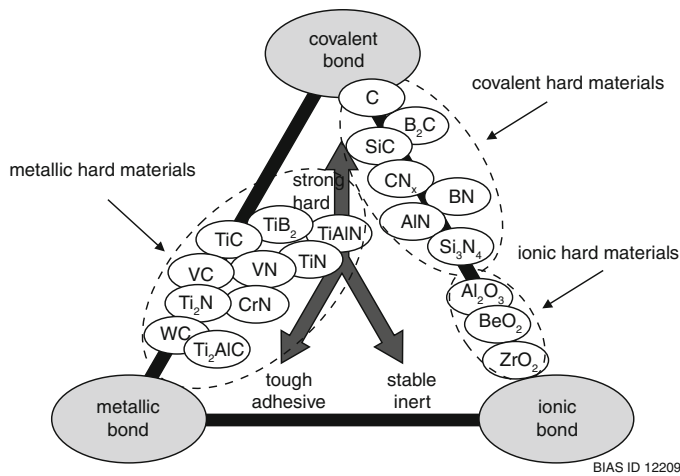


Fig. 7.76 Different hard materials and their characteristic bonding [39]

treated base material and the hardness of a widely used solid lubricant (MoS₂) are also listed in Fig. 7.75.

For single-layer coatings, the selected material must possess in addition to a high hardness, high adhesion to the substrate and a high chemical stability. Good adhesion can be achieved by metallic bonds, which have the highest adhesion. The hardness of a material usually increases with the proportion of covalent bonds. In

contrast with this, chemical stability, which prevents interaction with the surrounding lubricant, is achieved by ionic bonds. These are reasons why TiN has established itself as one of the first successful hard material coatings which combines all three required properties.

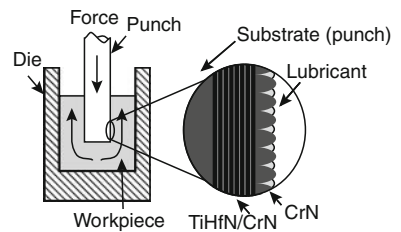
The posteriorly developed layers consist mostly of nitrides and carbides of titanium and chromium and the large class of carbon layers. Examples are TiN, TiCN, TiAlN, CrN, a-C:H and carbon in the shape of diamonds. Although multi-layer coatings are becoming more and more important, the selection of single hard material coatings as a final protective coating still plays a major role.

7.5.3.2 Surface Structure

Due to the high pressures at the interlayer between tool and workpiece during forming processes, the tool surface demands high abrasion and adhesion resistance combined with high hardness and a low Young's modulus. It is evident that a structured surface has to resist the stresses during the forming process, without changing its properties. A coating, such as the lotus effect, therefore only has a small chance of success. However, several groups deal with the structuring of the surface, for example to reduce the friction and to minimize the use of lubricant. Another approach is the deliberate alteration of the surface roughness and the associated change of the tangential friction force. A frequently used approach is lubrication pockets. On the one hand these allow for a more homogeneous distribution, and on the other hand a reduction in the use of lubricant. A layer combination as used to minimize the punch force in backward extrusion is shown in Fig. 7.77. Located above the base material, a nanolaminated layer system consisting of TiHfN and CrN is deposited on a tungsten carbide tool. In the right part of the picture a schematic structure of the uppermost layer is shown, used for the reduction of friction by the use of ester as a lubricant. Bobzin et al., demonstrated that with the help of this layer, the punch force, in comparison with high-speed steel, can be considerably reduced [5].

The application of self-lubricating materials is conducted in two ways. A widely used method is to undermine dimples in an existing hard material layer and the subsequent deposition of a solid lubricant. These kind of self-lubricating coatings combine the high hardness of the hard material layer and the good sliding properties of the soft coating.

Fig. 7.77 Textured coating for minimizing the use of lubricants [5]



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7.5.3.3 Lubrication

As shown in the previous section, hard coatings are widely used to improve the performance of forming tools. But these coatings are not necessarily useful for low friction and do not provide any protection for the opposing surface. As shown in section 2.2.2, the wear of the opposing surface is given by the choice of material, their contact conditions and the resulting friction. For this reason, in some cases, materials are used which protect the opposite surface by reducing the friction. It can be distinguished between two friction-reducing concepts. First, the hard material coatings which even without lubrication have a low friction coefficient, and second self-lubricating coatings that form a solid or liquid protective layer between the tool and workpiece. The first group mainly includes hard amorphous and crystalline carbon layers. In dry loading the friction coefficient is typically 0.15. However, these coatings do not provide an efficient transfer film on the opposing surface. The group of self-lubricating coatings is dominated by MoS₂ and graphite based materials. They enable friction coefficients down to 0.01.

The development of MoS₂ started in the 1950s. MoS₂ consists of a lamella bond by weak van der Waals forces, like graphite. Hence it can act as a solid lubricant. But in the beginning there was still a rapid rise in friction through degradation by contact with humid air. The use of ultrahigh vacuums in 1995 resulted in a significant enhancement of lifetime through the reduction of oxygen during the coating process. One of the latest developments is the deposition of a titanium interlayer or doping with titanium. This led to an increased hardness (1,000–2,000 HV) and a low sensitivity to humidity of the deposited MoS₂ layer [81].

The application of self-lubricating materials is conducted in two ways. A widely used method is to undermine dimples in an existing hard material layer and the subsequent deposition of a solid lubricant. These kind of self-lubricating coatings combine the high hardness of the hard material layer and the good sliding properties of the soft coating.

The second approach is a complete coating of the tool by a self-lubricating material. A disadvantage is that these materials are normally very soft and therefore usually wear quickly. In particular, some self-lubricant layers are also exceptional against humidity. However, the properties of self-lubricating layers are continuously improving so that they are a good alternative to hard coatings.

7.5.3.4 Material Properties

Suitable materials for interlayers are the range of hard material coatings, but also soft and ductile materials such as ZrC. The main applications are for an increase in adhesion, the prevention of diffusion of impurity atoms into the tool, and a reduction in residual stresses. A typical example where all three demands are given for the interlayer is the CVD-deposition of diamond onto TiN (Fig. 7.78).

Due to the high reactivity of carbon during the CVD process, diamonds need interlayers to deposit on materials like steel. One of the possible choices of

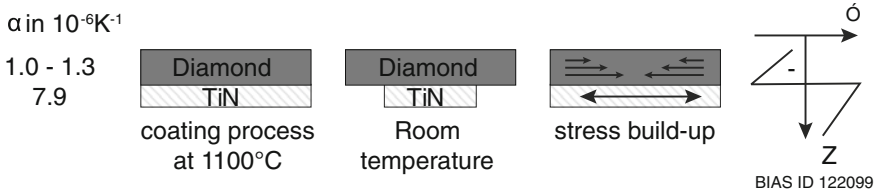


Fig. 7.78 Creation of residual stresses due to the high temperature during the coating process

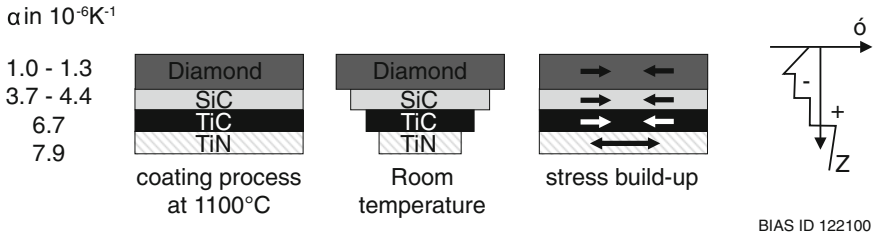


Fig. 7.79 Layer system to reduce the residual stress

interlayer that can withstand the temperature and reactivity of carbon is TiN. However, due to the the high temperature during the CVD process, the different thermal expansion coefficients of the substrate and the hard material coatings, internal stresses result after the device has cooled down. In Fig. 7.79, the different thermal expansions of the TiN-layer and the diamond coating are given. The deposition temperature during the CVD-process is about 1,100 °C. After the coating process is complete, the multilayer cools down to room temperature. For this reason a high transition between compressive stress and tensile stress is created between the coatings. The consequence may be that the layer breaks off or the adhesion is reduced. In addition, the internal stress can reduce the lifetime of the component by continuous usage. In Fig. 7.79 it is shown how the internal stress can be distributed by the interlayer, with materials which have a thermal expansion coefficient between TiN and diamond. In the example titanium carbide and silicon carbide are deposited on the inter-layer. This can avoid the abrupt transition of the expansion coefficients. Challenges in the use of inter-layers are the different process temperatures of the deposited coatings.

Another problem in addition to the different thermal expansion coefficients is the catalytic effect of ions in the preferential nucleation of graphite and the high solubility of carbon during the CVD-process. Therefore many studies on interlayer systems between the steel substrate and the diamond coating have been reported. The most common interlayer for diamonds are Cr-, Ti- and Zr-based nitrided and carbonized materials [77]. However, there are also Mo/Ag/Nb based multilayer systems presenting good adhesion and reduced stress.

7.5.3.5 Layer Architecture

In addition to the components, multilayers can also vary in terms of their architecture. An overview of different layer types is given in Fig. 7.80. The simplest form of coatings is the monolayer. Typical samples are the early hard coatings as shown in Fig. 7.77, and the self-lubricant coating MoS_2 . The advantages and disadvantages of these layers are discussed in the previous sections.

The next logical step is to deposit two layers instead of one, and to combine their properties. However, the idea of combining hard and soft materials to improve the properties such as hardness and stiffness is not an invention of modern times. One of the first references is in “The Iliad” by Homer, which describes Achilles’ shield as a laminate out of bronze/tin/gold/tin/bronze. Even prehistoric blades, often referred to as Damascus swords, are welded out of high carbon steel and low carbon steel. However, the depositing of combined hard and soft coatings with PVD and CVD processes still starts at the end of the 20th century. Especially for very hard carbon and boron nitride coatings, soft interlayers are used as bonds to the tool.

Interlayers between two different coatings can always be a weak point in the architecture of the component. Especially when stress peaks are generated at the transitions between the coatings, where it can result in a separation of the layers and a failure of the component. One possibility to avoid this is to model the transition between the two layers as a gradient. By choosing suitable materials, the hardness can be increased continuously from the workpiece to the surface without producing a sharp transition and a weak point in the coating system. However, the use of graded coatings is only partly possible because on the one hand, most materials cannot be converted into each other and on the other hand the process conditions for the production of the layers are too different.

The sequential deposition of two different coatings is commonly referred to as multilayer. Despite the much higher complexity, these types of coating are

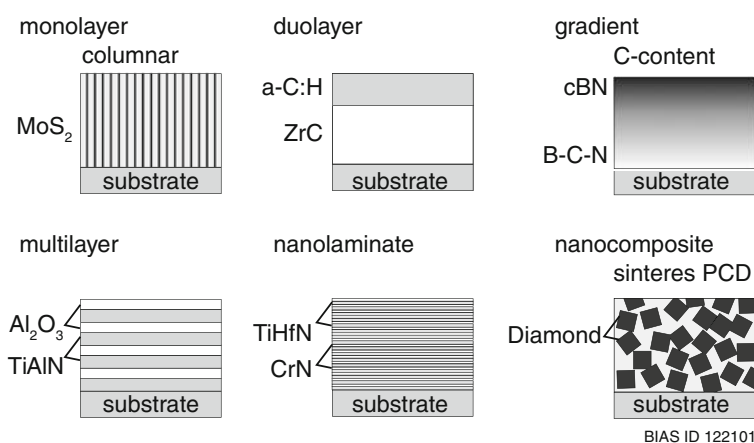


Fig. 7.80 Different layer architectures

becoming more and more used. This can be explained by the improvement in the mechanical, thermal and physical properties compared to common single-layer coatings. Advantages are, in addition to an increase in layer hardness with a simultaneous increase in fracture strength as with hard-soft combinations, a prevention of crack growth through the interfaces of the layer. Due to the rapid sequence of thin layers, and the resulting higher proportion of the nucleation phase, the columnar growth of PVD coating in multilayer environments decreases. The coatings become more compact and smoother. This results in a better diffusion barrier. Furthermore, the internal stress can be reduced by the repeated grain growth at the beginning of a new deposition.

Reduction of the layer thickness increases the number of interlayers and increases the influence of nanocrystalline structures. The resulting layer structure is called nanolaminate. This can lead to higher surface energies, better barrier properties and thus to good oxidation resistance.

In addition to linearly-built coating systems, there are also some other layer constructions which promise to improve the component properties. Here only the nanocomposite is named, as shown in Fig. 7.80. Due to the large proportion of interfaces in a nanocomposite, high hardness values can be achieved.

7.5.3.6 Coatability

In most cases, a coating is applied to an existing tool to improve its features. Thereby the material of the tool is fixed and the coating must be adapted to it. Suitable substrates are therefore the entire range of steels, hard metals and ceramics. But not every coating can be deposited on any material. The key factors are how the base material behaves at high temperatures and its chemical resistance against oxygen, nitrogen, hydrogen and carbon. For metal forming and cutting tools, polycrystalline CVD diamond coatings possess some of the preferred properties like hardness and wear resistance [107]. However, beside the internal stress through the high deposition temperature of about 900–1,000 °C, the structure and geometry of tool steels is heavily influenced. For this reason, usually hardening and tempering under vacuum conditions after the process is necessary as well as the post processing of the contours.

Beside the effect of temperature, the tool can also be damaged by diffusion of atoms and molecules during the coating process, or through chemical reactions with the precursor or side products. As an example, the deposition of diamond coatings on steel without an interlayer is not possible, since the solubility of carbon in steel at the deposition temperature is high. But even the deposition of diamond on hard metals like cemented tungsten carbide (CC) cannot be executed without special surface treatment. The catalyzing influence of cobalt on the diamond growth and the large mismatch between the thermal expansion coefficients is responsible for this. Without an adjusted surface treatment the cobalt content leads to void formation at the interface, the presence of non-diamond material at the interface, high residual stresses and therefore a lack of good adhesion. Various

Fig. 7.81 Typical cross-section of an etched tungsten carbide (CC) sample with a diamond coating [12]

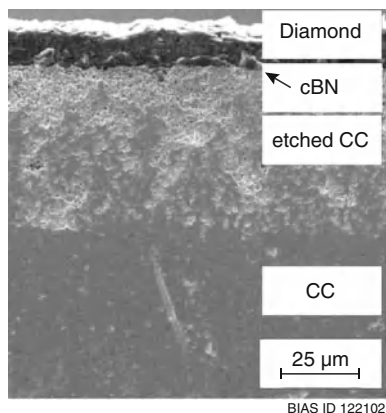
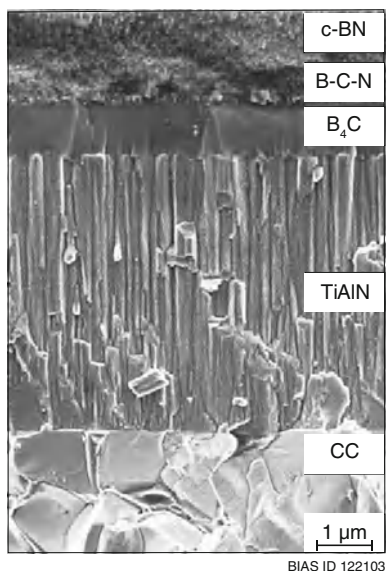


Fig. 7.82 Boron nitride-based multilayer coating [93]



pretreatments techniques are surface etching with aqua regia, Murakami treatment and subsequent ultrasonic micro scratching with a fine diamond suspension. An example construction of a diamond coating with a cBN interlayer and an etched tungsten carbide substrate is shown in Fig. 7.81. It can be easily identified that the etching process removes the cobalt to a depth of up to 40 microns.

The main advantages of PVD coatings are the low temperature and the wide range of usable materials. The substrate temperature is usually in the range of 300–600 °C. At this temperature there is no distortion or structural change of the base material. But by the absence of diffusion processes throughout this temperature, the adherence of the coatings is limited. An example of a cubic boron nitride-based coating system deposited by PVD processes is given in Fig. 7.82.

In contrast with the diamond multilayer, a TiAlN layer protects the tungsten carbide substrate from a reaction with the necessary carbon during the following coatings. Furthermore, gradient and adhesion layers are deposited to improve the coating-system's properties.

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Chapter 8

Machines and Handling

Bernd Kuhfuss

Variables

a	Distance (mm)
v	Velocity (mm/s)
d _S	Ball diameter (mm)
d _w	Wire diameter (mm)
F	Force (N)
p	Pressure (psi)
U	Electrical potential (V)
T	Temperature (°C)

8.1 Presses

Hendrik Schulze Niehoff

8.1.1 Classification

Most micro metal forming processes need a linear, mechanical movement in order to initiate and conduct the forming process. This applies especially to most bulk metal forming processes, such as upsetting and cold forging, described in [Chap. 4](#), and to sheet metal forming processes, such as deep drawing, stretch drawing, bending, flange forming, piercing and blanking, described in [Chap. 5](#). Nevertheless, there are some processes with rotational movement, such as rotary swaging, or other incremental forming processes. Furthermore, there are micro forming processes, which

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do not need any external mechanical movement at all. They obtain the forming energy, for example, by laser induced shock waves, such as in laser shock forming.

The linear movement is provided by forming machines, called presses. The requirements and properties of presses for use in micro forming processes will be the subject of the following subchapters. Presses are traditionally classified into work-linked, force-linked and travel-linked presses. Additionally, in the recent past, presses have appeared which combine several mechanisms of action. They can, therefore, be called hybrid presses.

8.1.2 Work-Linked Presses

Work-linked presses are hammer and screw presses. They are cheap and can supply high force. Therefore, they are especially used for bulk forming processes needing high force. The work of hammers is linked to the potential energy of the ram. The work of screw presses is linked to the kinetic energy of the rotation of the fly wheel. Due to very poor accuracy, work-linked presses are not suitable for micro forming processes.

8.1.3 Force-Linked Presses

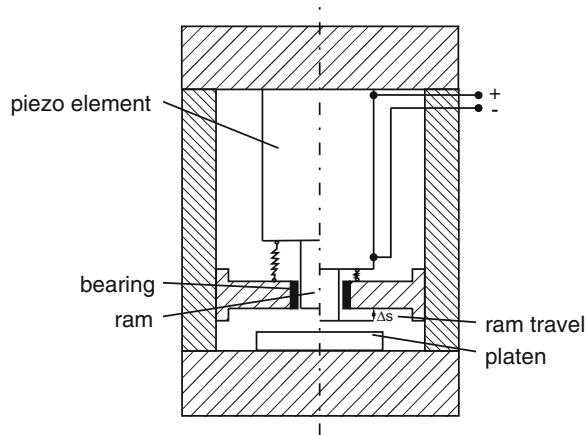
Force-linked presses are characterized by the fact that the maximum force is available at any position of the ram. This is especially useful for bulk forming processes. The ram force can be generated either hydraulically, pneumatically, piezo-electrically or by servo drives.

Hydraulic presses act with the pressure of a fluid, which is transformed to mechanical movement by a cylinder–piston system. Since the travel reverse is not defined by a certain pressure, the dead center can be set at any position of the ram. But for processes which require high positioning accuracy, a mechanical stroke limitation is necessary. Furthermore, the stroke rate is limited to 400 strokes/min by the inertia of the fluid. The use of hydraulic presses in micro forming processes is, as yet, unknown.

Pneumatic presses act in a similar way to hydraulic presses, whereby the fluid is replaced by air. The lower air inertia allows higher stroke rates, but the higher compressibility causes a lower stiffness, which can result in increased vibration at the lower dead center. Thus, pneumatic presses are rather more suitable for assembly jobs than for micro forming processes [20].

Piezo presses are currently only an object of scientific research and not available in the market, although piezo actuators exist in a wide range of applications. They use the piezo effect, where the piezo elements expand proportionally to the applied electric voltage and thus generate the ram movement, see Fig. 8.1. The advantage of this drive is comparable high acceleration and high accuracy

Fig. 8.1 Principle of piezo press; *left: upper dead center; right: lower dead center*



simultaneously. The disadvantage is that the total height of the actuator has to be 1,000 times the stroke height. An important milestone was reached by Hess [12], who built a fully functional piezo press. He combined the piezo actuator with a hydraulic stroke enhancer in order to increase the stroke length from 120 μm to 2.85 mm. At the same time, the maximum force was reduced by 90 % to 350 N. The intended stroke rate of 3,000 strokes/min could not be reached due to lack of control performance. The measured stroke rate is 420 strokes/min at a stroke height of 2.2 mm. The free travel-time control of the piezo actuator was of special benefit in blanking applications, where the share of the flush-cut could be increased from 70 to 100 % by changing the travel-time characteristic from sinus to rectangular. Another piezo press was built by Sautome [19] for micro backward extrusion processes.

Servo presses have had more and more impact recently. Their design is quite similar to the crank press, but they have no flywheel and the electric motor is replaced by a torque motor with a very high moment of force. This allows it to freely control the press in its travel-time characteristics. Bihler (Germany) uses its servo press SP 500 in combination with its NC production and assembly system BIMERIC, see Fig. 8.2. This all-in-one production system can be equipped with a wide range of other NC units such as feeding units, stamping and forming units, welding units, thread forming units, screw insertion units or pick and place units. The servo press, SP 500, reaches a repeat accuracy of 10 μm at a maximum stroke rate of 400 strokes/min. The maximum stroke height is 60 mm, the maximum force is 500 kN.

8.1.4 Travel-Linked Presses

Travel-linked presses are driven by crankshafts, as in eccentric, toggle or crank presses. The travel-time characteristic is directly linked to the mechanical parts

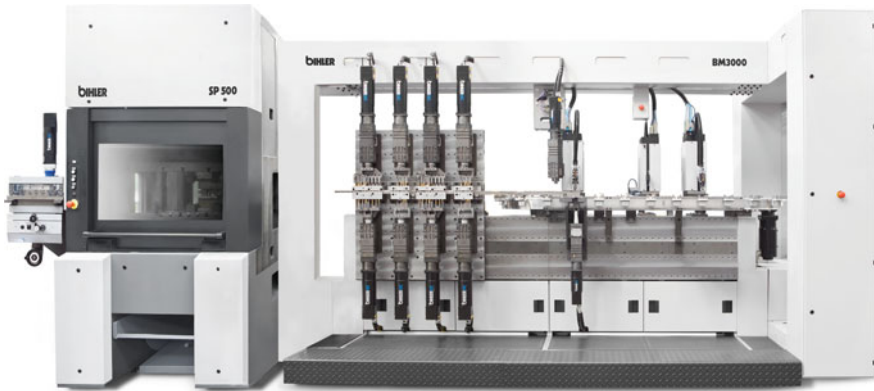


Fig. 8.2 Servo press SP 500 as part of the NC production and assembly system BIMERRIC, Bihler GmbH & Co. KG (Germany)

and, thus, is basically sinusoidal, see Fig. 8.3a. Slightly different characteristics can be gained by a modified crank mechanism, as in toggle presses. This applies especially for deep drawing processes, where the approach to the lower dead center can be realized more gently. The link of the ram travel to the mechanical parts allows the ram position to be adjusted at the lower dead center very accurately, limited only by the slackness of the mechanical parts, see Fig. 8.3b.

If the press design is rather tight and small, then slackness, deflection and torsion are significantly reduced, which leads to an increase in stiffness and accuracy (below 30 μm). The lower accelerated masses then allow very high stroke rates between 300 and 2,000, occasionally up to 4,000 strokes/min. This qualifies crank presses for micro forming processes. Many different kinds are available on the market. The fastest known are Omega F1, from Yamada Dobby (Japan), and BSTA 200, from Bruderer AG (Switzerland), see Fig. 8.4.

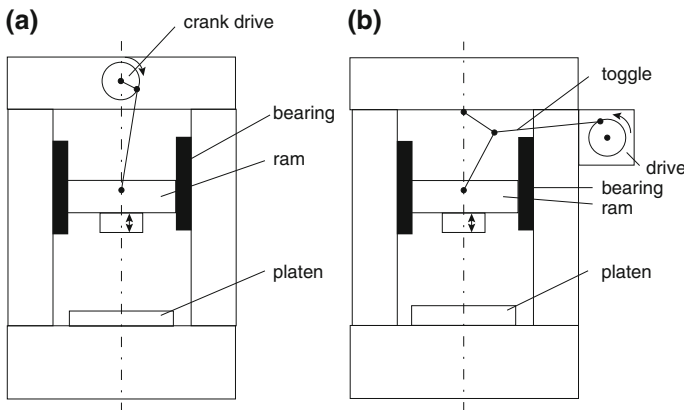


Fig. 8.3 Principle of **a** crank and **b** toggle press

Fig. 8.4 Bruderer high performance stamping press BSTA 200, Bruderer AG (Switzerland)



8.1.5 Hybrid Presses

Hybrid presses combine several mechanisms of action, for example, a crank press with a hydraulic auxiliary drive in order to combine the high stroke rate of the crank press with the free travel-time characteristic of the hydraulic press. Another type of hybrid press is the linear motor press since it can be seen as force-linked, due to the electric linear motor, but also as work-linked, due to the kinetic energy of the ram which is put into deformation work. The principle is based on an uncoiled electric drive, where a stationary electric motor induces a Lorentz force on permanent magnets, which are fixed to the ram and thus lead to a linear movement, see Fig. 8.5. There are further designs possible where the electric linear motor is the moved part or asymmetric designs where only one electric linear motor is used. Furthermore there exist different types of electric linear motors, such as step motors, DC-motors, synchronous and asynchronous motors, whereby the linear synchronous motor has the highest force related to the moved mass and allows therefore the highest acceleration. A relatively rare electric linear motor is the solenoid motor, where the electric motor is not a plate but tube halves with circular slots.

The accurate control of the linear motor in combination with few moving parts, allows a very high accuracy up to 1 μm . If the mass of the moving parts is kept low, then high accelerations and thus high stroke rates up to 1,000 and occasionally up to 3,000 strokes/min can be reached. The travel-time characteristic of linear motors is free and the maximum force is available at any position of the ram. The disadvantage of linear motors is that the maximum force is limited to 20 kN per motor, but this means no limitation for micro forming applications. The unique combination of very high accuracy, high stroke rates and a free travel-time

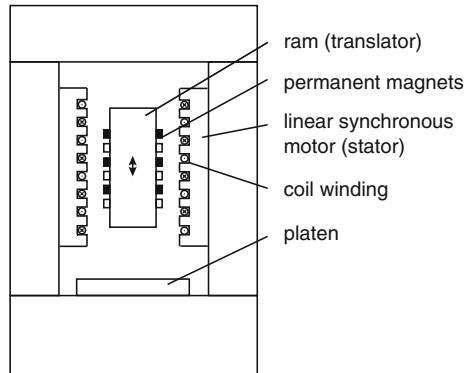


Fig. 8.5 Principle of linear motor press

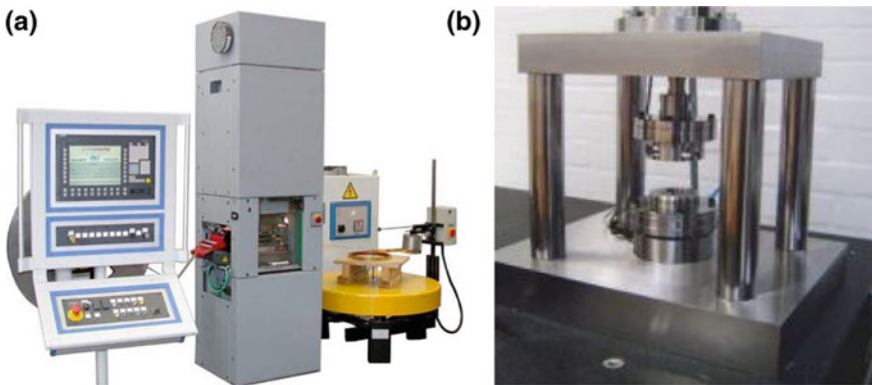


Fig. 8.6 **a** Linear motor press stanzrapid (PTU, Darmstadt), **b** Solenoid press (Technical University of Denmark)

characteristic makes the linear drive press the most applicable principle for micro forming processes.

Although they are rarely available on the market, several linear motor presses have been built. Schepp [20] has built the so called “Stanzrapid” (Rapidstamper) with two opposing linear motors. The press is commercially available from Schuler AG (Germany). The use of plain bearings leads to an accuracy of $5.6\ \mu\text{m}$ at a stroke rate of 1,200 strokes/min. The stroke height is infinitely variable up to 80 mm, whereby ram velocities of 2 m/s can be reached through a maximum acceleration of $110\ \text{m/s}^2$, see Fig. 8.6a. Another linear drive press has been built by Paldan [17] at the Technical University of Denmark. He used a solenoid motor, allowing very simple and cheap press design. A positioning accuracy of $1\ \mu\text{m}$ could be reached at a stroke rate of 300 strokes/min. This laboratory press has a stroke height of 100 mm, see Fig. 8.6b.

The most recent work is the development of the “MUM”, a linear motor press especially engineered for micro forming processes. The design and properties of this press are described in the next chapter.

8.1.6 Engineering of a Press

The engineering of a press with special focus on micro forming processes is described on the basis of the MUM, a linear motor press which was developed at BIAS at the University of Bremen [21, 22].

Requirements

The press is focused on applications of deep drawing processes for metal parts of 0.5–10 mm in diameter. Therefore, a repeat accuracy of 3 μm or less is needed to meet the required tolerances of the parts. The press force has not only to exceed the punch force for such parts, which is far below 3 kN, but has also to supply the high acceleration, which is needed to reach stroke rates of 1,000 strokes/min to be competitive with commercial crank presses. In order to gain high flexibility with regard to different forming processes, a very high stroke height of 200 mm was aimed for. Furthermore, it was necessary to have a variable ram travel versus time characteristic in order to allow the examination of the advanced process design. Additionally, force controlled travel should be possible, for example to reduce the blank holder force during the deep drawing process and thus to increase the limit drawing ratio. The force control yields to the demand that the force has to be measured with an uncertainty smaller than 0.05 N. To control the blank holder independently from the punch, a second axis will be needed. All in all, a fast, double-axis, high precision press is demanded.

Drive

Many presses are driven by a crank motion, especially if a high output quantity is demanded. But the crank motion is not suitable since the punch travel is not a freely controllable characteristic and force control is not possible. A hydraulic or pneumatic drive would meet these requirements, but they are rather slow. A piezo drive is fast as well as free in force and motion control, but its punch travel height is limited to a few millimeters. Thus, an electrical linear motor was chosen. It is fast, free in force and motion control and can make high strokes. Another advantage of the linear motor is its high precision. In order to reach high acceleration, the ratio of moved mass to force of the linear motor is decisive. It was found that this ratio increases in larger motors. On the other hand, the motor should be rather small to reduce disturbances, such as heat evolution, attraction forces and snap in forces. It was also calculated that higher accelerations can be reached if the magnets instead of the stator are moved. This also has the advantage that the power and coolant supply do not need to be carried with the moved part.

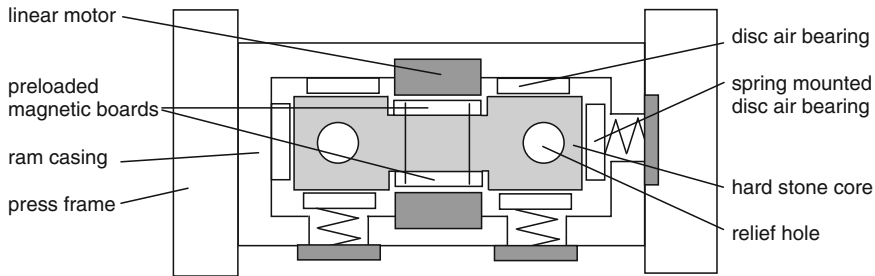


Fig. 8.7 Cross section of the ram unit, ram casing and press frame

Two linear motors per axis were installed in opposite orientations (Fig. 8.7), so that the attraction forces between the linear motor and the magnetic board of 15.4 kN nearly cancel each other out. That makes the use of air bearings possible as they can bear only small forces.

Bearings

The proper design of the bearings is essential to the press, since the bearings can produce forces which disturb the measurement and thus the control of process forces. Former tests showed that, for example, ball bearings are not suitable due to disturbing forces, which are at least in the order of 1 N. Aerodynamic guide slide bearings would have similar effects due to the stick-slip effects, especially for the reversing movement. A hydrostatic bearing would need a high flow rate for the fluent since the arrangement of movement is vertical. The contact-free magnetic bearing is an interesting solution for the press, but it is not available on the market and the design is very complex.

Thus, the decision was made for aerostatic bearings, which are contact-free and non-wearing. However, they are very heat sensitive due to the small air gaps of only 5 μm but can be designed as flat discs or as pillars. Figure 8.8 shows the cross section of a two-pillar air bearing design in comparison to a disc air bearing design, subjected to a change in temperature of 1 K. The change in shape of the aluminum ram leads to a reduction of the air gap from 5 to 1.5 μm , which is in the order of the geometric accuracy of the pillar, so that locking could occur. In contradiction to this, stands the use of disc air bearings, which are cheaper and heat tolerant, since the change in shape can be compensated by disc springs.

Another advantage of the disc air bearing is the higher load and stiffness compared to pillar air bearings. While the pillar air bearings would have had a nominal load of 4 kN, the disc air bearings have a nominal load of 6.8 kN and a stiffness of 800 N/ μm . The springs for temperature compensation would, of course, reduce this stiffness significantly and vibration could occur in fast reversing applications. It is, therefore, a hydraulic linear brake integrated into the bearing system, which can lock the bearings and deactivate the effect of the springs if needed. If, in this operational mode, changes in temperature occur, only a short unlocking is needed and the bearings are again self-adjusted. For the worst case of

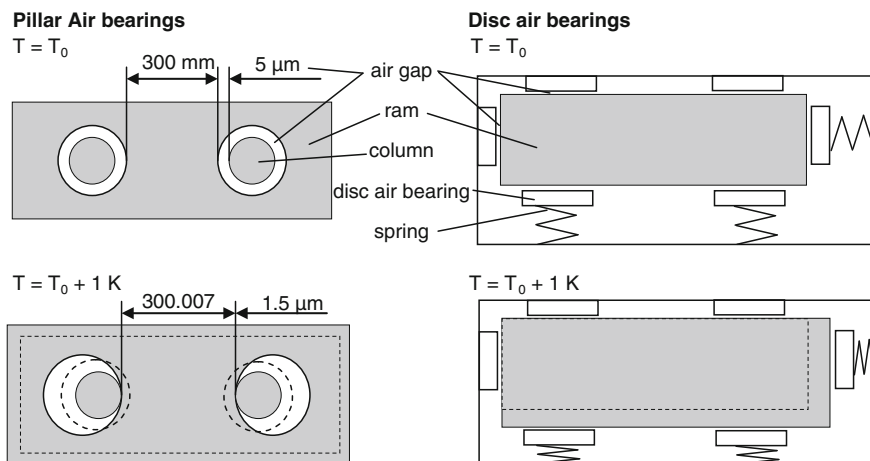


Fig. 8.8 Effect of temperature change of 1 K on air bearing with pillars and with discs

unavoidable contact, the disc surface is armed with a dry-running coating which lasts for more than 5 km.

Ram

The use of disc air bearings requires very good planarity and parallelism of the bearing surface. Thus, a ram core of hard stone was taken, enabling surfaces with a planarity of $1\text{ }\mu\text{m}$ of 1,000 mm in length. The naturally grown stone relieved all residual stresses, which guarantees the absolute absence of shape distortion and a lifetime of parallelism. Furthermore, the hard stone has the same density as aluminum, which makes it light, but its thermal expansion coefficient is only a third, which reduces temperature based changes in shape.

The magnet boards on both sides of the ram have to be connected directly to each other through holes within the hard stone, in order to preload the hard stone with pressure, so that no tensile stresses are acting on the hard stone, if the attraction forces start to work, since the hard stone cannot bear tensile stresses. The same strategy is used for the upper and lower ram plates, which are also preloaded through tie rods. The plates also act as a stop unit for the security dampers, which come into operation if the electronic security devices fail. But, before they act, linear brakes are designed to lock the ram if power failure occurs, the emergency button is pressed or if an electronic part of the machine control fails. Furthermore, a locking pin was designed to achieve positive controlled for tool changes and adjustments, Fig. 8.9.

Entire construction

The press is based on a hard stone plate, which is anchored to the ground by shock absorbers in order to reduce vibrations. The extreme planar surface of the ground

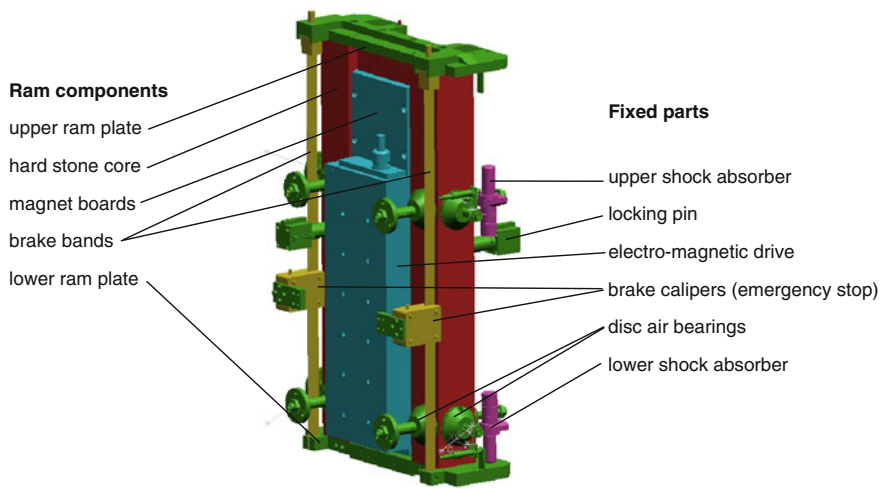


Fig. 8.9 Ram unit and connected components

plate is the reference for the two columns, which are adjusted orthogonal to the ground plate, and all other components. The columns are a welded construction, which is filled with polymeric concrete to increase the damping effect. Two ram cases are installed between the columns, which contain the linear drives, the air bearings, the brakes, the locking pins, the shock absorber and the rams. The ram cases can be dismantled and assembled without further adjustment of the machine. The workbench is not only a fixing plate for tools, but also a stiffness component. It stands on leveling feet and is fixed to the columns. The height of the workbench

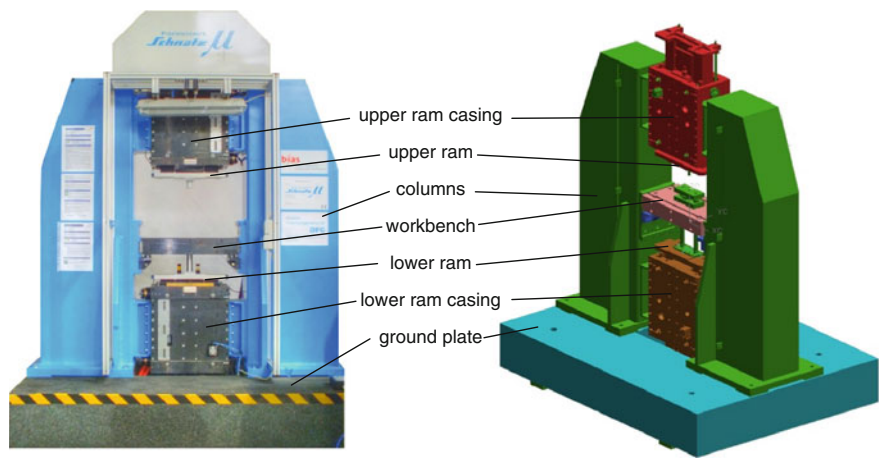


Fig. 8.10 Entire construction of the MUM

Table 8.1 Properties of known presses

Name of press	MUM	Stanzipap	Solenoid press	Piezo press	Omega F1	BSTA 200	SP 500
Manufacturer	BIAS	Schuler AG	IPU	ZFS	Yamada	Bruderer	Böhler
Number of axes	2	1	1	1	1	1	2
Drive	Electro-magnetic	Electro-magnetic	Electro-magnetic	Piezo-electric	Crank motion	Crank motion	Servo motor
Control	By force or position	By force or position	By position	By force or position	Fixed	Fixed	By position
Maximum force	15 kN	20 kN	5.5 kN	0.38 kN	100 kN	200 kN	500 kN
Positioning error	3 µm	5.6 µm	1 µm	—	<5 µm	<30 µm	10 µm
Maximum velocity	3 m/s	2 m/s	2.5 m/s	—	13 m/s	—	—
Maximum acceleration	10 g	11 g	40 g	10 g	—	—	—
Stroke rate	1,250/min at 1 mm stroke height*	1,200/min	300/min	420/min	4,000/min at 6 mm stroke height	2,000/min at 8 mm stroke height	400/min
Stroke height	200 mm	80 mm	100 mm	2.3 mm	20 mm	76 mm	60 mm
Installation space (H × W × D)	200–400 × 600 × 300 mm	230 × 230 mm	200 × 300 × 150 mm	—	235 mm	260 × 410 × 400	311 × 1,000 × 560 mm
Bearings	Air	Plain	Rolling	—	Plain	—	Plain

*2,900/min with higher positioning error

can be varied in three positions, which differ by about 100 mm, so that the upper or the lower installation space can be changed from 200 to 300 or 400 mm in height respectively. Three holes in the workbench allow the interaction of the rams directly (through the middle hole) or the transfer of one ram movement in either the upper or the lower installation spaces (through the two outer holes). The width of the installation space is 600 mm, which provides good accessibility, Fig. 8.10.

Properties

The developed press has two axes, which can interact and allow new processes. It is very dynamic and can reach 2,900 strokes/min. The linear motor combined with the air bearings ensures the absence of friction and clearance. Therefore, a high positioning accuracy of less than 3 μm and a repeat positioning error of less than 1 μm at 1,250 strokes/min have been achieved. The combination of speed, accuracy and flexibility makes the MUM unique from other described presses, see Table 8.1.

8.2 Special Purpose Machines

Volker Piwek and Bernd Kuhfuss

8.2.1 Design of Rotary Swaging Machines

Generally there are three different types of machine for metal forming by rotary swaging. For noncircular geometry it is necessary that the rotary shaft is locked and that the tools do not perform any rotation. This can be achieved by the application of an external rotor or expeller configured machine with a locked rotary shaft. On the other hand, rotation-symmetric profiles can be manufactured on internally driven as well as both the other types of machine (Fig. 8.11).

The basic kinematics of the drive system for the tools depends on the size of the machines, and this size is a function of the dimensions of the machined cross-sections with respect to the final geometry to be achieved. Due to limited constructional space, currently small machines suited for micro forming are only designed with an internal rotor. On the other hand, principle process variants—infed and plunge methods—are realized with small rotary swaging machines.

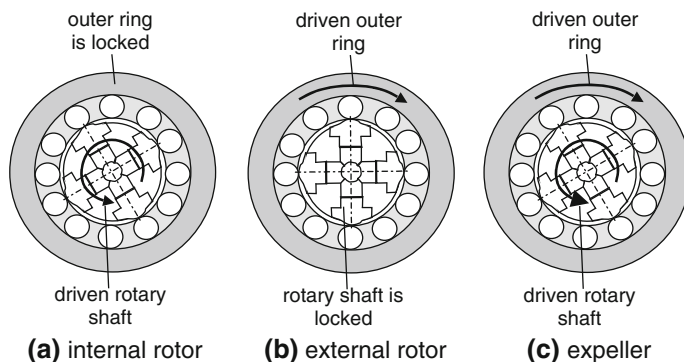


Fig. 8.11 Types of machine (schematic)

8.2.2 Extensions and Modifications of Small Machines Principally for Micro Forming

8.2.2.1 Synchronization of Workpiece Rotation

To make noncircular geometries producible in the micro sphere the workpiece can also be synchronously rotated with the speed frequency of the rotary shaft. Such a machine kinematic has been realized as a prototype and proved within Collaborative Research Centre Micro Cold Forming (Fig. 8.12).

The rotary motion is directly deducted from the swaging head by a spindle which is connected to the synchronization gear drive. There the rotation is transformed with the same reduction gear ratio that is realized in the swaging head. The synchronized rotation between the tools and the workpiece is coupled into a hub of an axial relocatable insert shaft, which is carrying the chuck for the

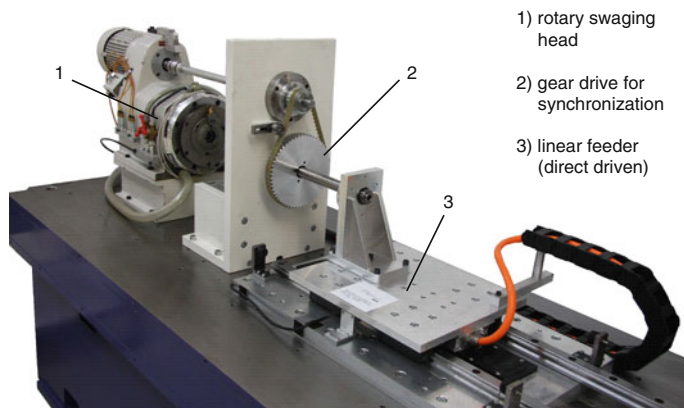


Fig. 8.12 Experimental station with synchronization of workpiece rotation and linear feed drive

workpiece at its front end. The axial feed with regard to the axial positioning of the workpiece is induced by a linear direct drive. The table supports the insert shaft at its rear end.

8.2.2.2 Axial Workpiece Feed Device

Forming in the micro domain by rotary swaging requires a higher accuracy on the axial positioning of the workpiece due to smaller tolerances. Otherwise the acting process forces on the feed unit are comparatively small, which enables the substitution of the conventional hydraulic linear axis by other active principles. The integrated direct driven feed unit of the prototype features high positioning accuracy as well as high dynamics (Fig. 8.12). Another design of the axial feed unit which is not exclusively used for micro applications is a continuous friction based system (Fig. 8.13). In particular, this kind of feed drive is applied when forming endless wire material. Normally such a system is supplemented by two coilers.

8.2.2.3 Radial Feed of the Tools

In plunge rotary swaging operations the radial feed motion of the forming tools requires increased positional accuracy. Conventional machines are working basically with the hydraulic linear axis for the adjustment of the crank, that moves the wedges into the gap between the base jaw and the forming tools. In micro process machines this axis has been substituted by an electromechanical solution, consisting of a servo drive and a ball screw to generate the feed motion of the crank (Fig. 8.13). Hereby a high positioning accuracy can be achieved according to the resolution of the encoder. Furthermore it becomes feasible to realize defined velocity–time profiles for the radial tool motion.

8.2.3 Development of the Potential of Dry Processing

Forming processes by rotary swaging are predominantly realized by using a cooling lubricant. Primarily this fluid enables the friction conditions in the contact zone between the tools and the workpiece to be influenced. It is particularly used in the production of macro components to flush the roughened reduction zone of the tools which is loaded-up with abrasion from the workpiece surface. Due to the significantly smaller process forces in the field of micro production it becomes feasible to avoid cooling lubricant, either partially or completely. The above-mentioned flushing function of the cooling lubricant can be dispensed with if the reduction zone is designed with a flat surface. For dry processing there are two essential requirements. On the other hand, cooling lubricant is also used to

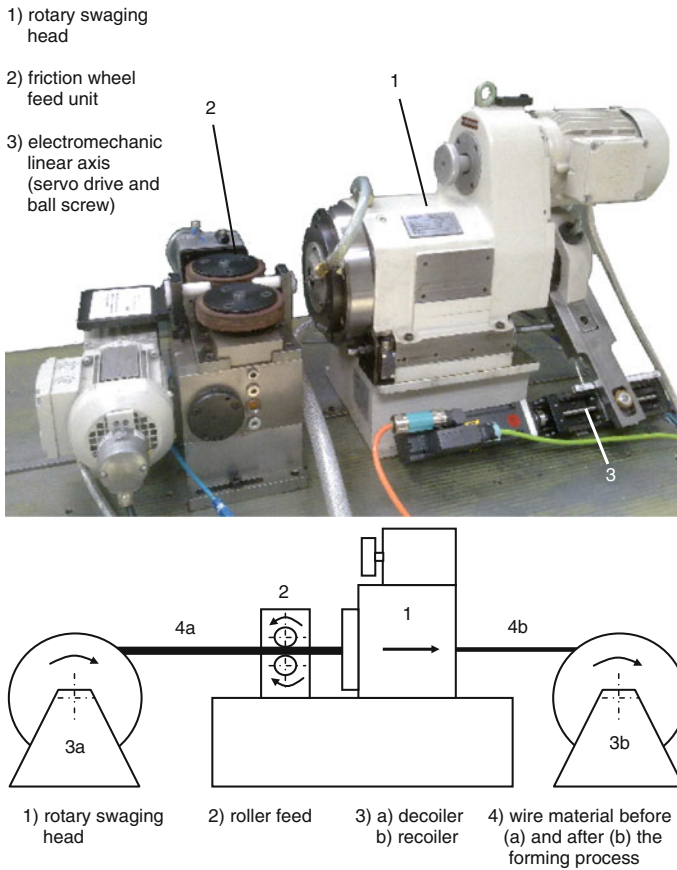


Fig. 8.13 Rotary swaging unit with a roller feed and an electro-mechanic linear axis (*top*), configuration to process wires consisting of two coilers (*bottom*)

lubricate all contact zones on surfaces that perform relative movement within the rotary swaging head. These conditions have to be maintained by dry processing, and adequate methods are necessary to encapsulate the forming zone from the area where the oscillating movement of the tools is generated to prevent the intrusion of the cooling lubricant into the contact area between the tools and the workpiece. This requires that the jaw disk has to be protected on the rear side from intrusion of the cooling lubricant, and furthermore no passage of the medium should occur in the radial guide slots between the frontal buffer plate and the tools. In the following two drawings, design features for the internal enclosure of the forming zone within the rotary swaging head are illustrated.

Figure 8.14 shows schematically a longitudinal section through the rotary swaging unit and the fitting conditions of the jaw disk, the wedge adapter and the swaging shaft. In order to prevent cooling lubricant from intruding into the forming zone, the adjustable gap (15–45 mm) between the rear face side of the stationary jaw

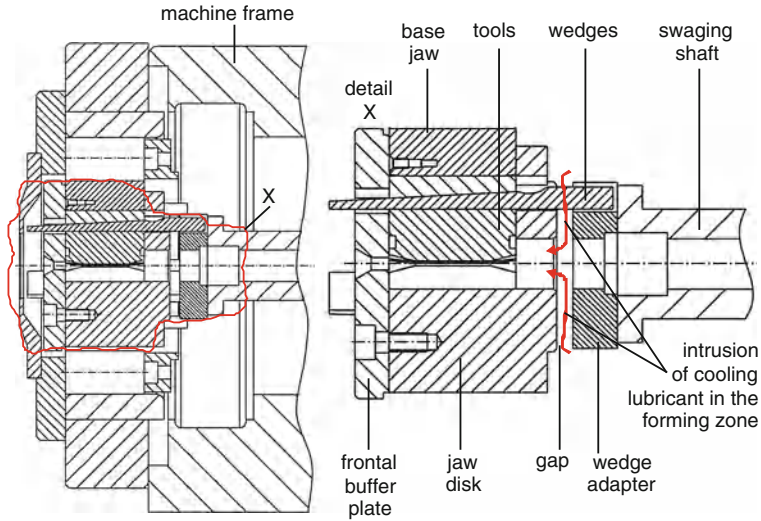


Fig. 8.14 Longitudinal section through the rotary swaging head

disk and the axial relocatable wedge adapter has to be closed. The design solution provides a flange with a frontal integrated o-ring as a seal against the jaw disk at the end of a tube, which is plugged into the hollow swaging shaft (Fig. 8.15). This tube rotates together with the swaging shaft and is supported outside the swaging unit by a ball bearing and an angle bracket on the machine base.

In the front area, seals are necessary between the tools and the radial guide slots in the jaw disk with respect to the frontal buffer plate. Figure 8.16 shows possible seals between the jaw disk and the frontal buffer plate as a modification of a tool segment for dry processing. In that case the enclosure is achieved by a slot and a seal in the bearing surface of the tool.

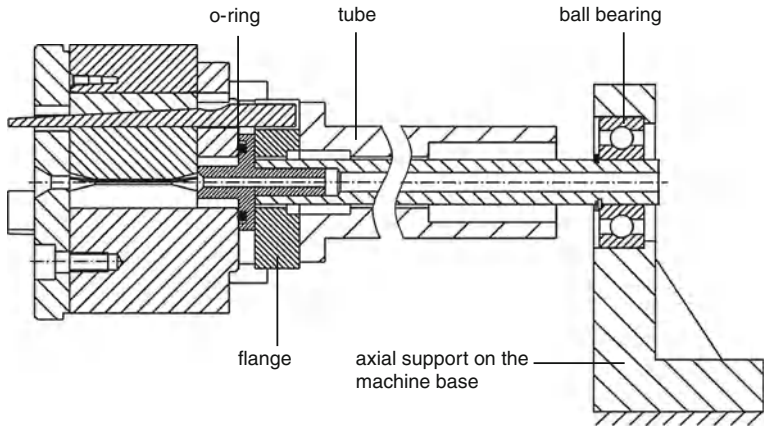


Fig. 8.15 Design solution for an enclosure of the forming zone in the rear area

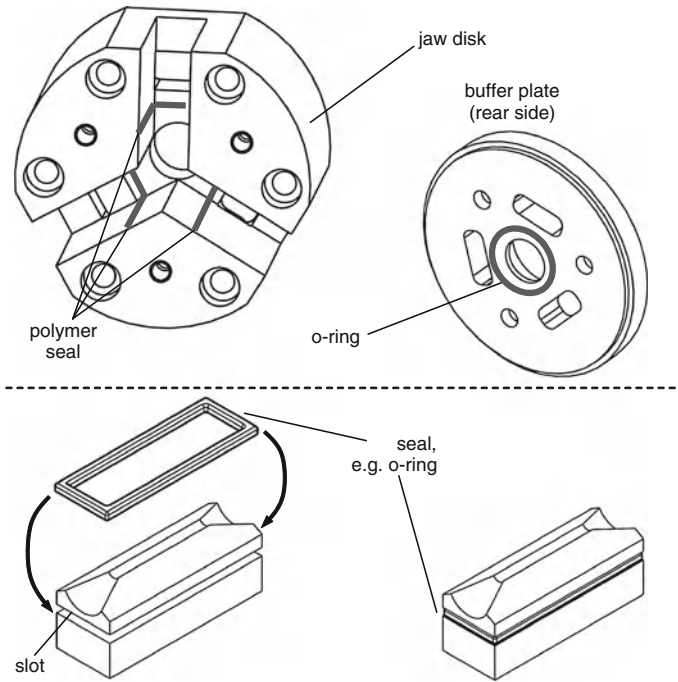


Fig. 8.16 Modification of jaw disk, frontal buffer plate and tools

8.2.4 Modification of Tools for Micro Forming

8.2.4.1 Adaption of Tool Osculation

Due to unfavorable contact conditions a reduced resulting osculation between the tools and the workpiece, especially using smaller machines with fewer tools in the micro area, a decreased roundness respectively cylindricity is expected (Fig. 8.17).

One method to compensate for this effect is an aligned mechanical release of the tools in the area facing the contact zone. The purpose is to increase the osculation with respect to the load during the contact between the tools and the surface of the workpiece. Figure 8.18 provides an overview where such mechanical releases can be located in areas of the tool's cross-section.

In terms of the high surface quality of the workpiece, it is practical to keep the contact area continuous (designs b and d).

For the design with lateral release (b), a finite element calculation has been conducted. Figure 8.19 shows the results of the comparison stress σ_c and the overall deformation.

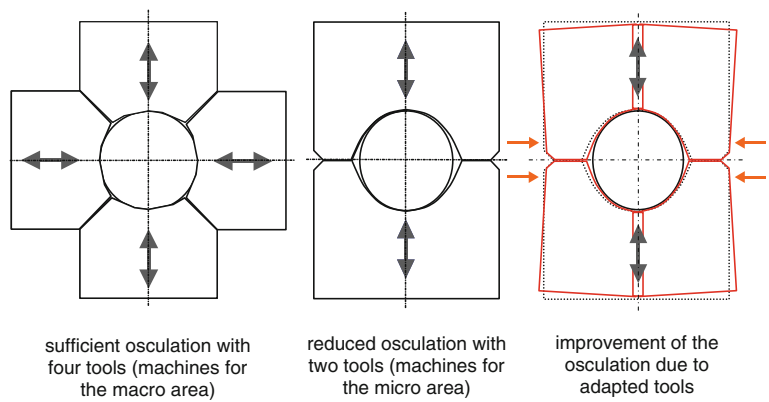


Fig. 8.17 Osculation between the tools and the workpiece

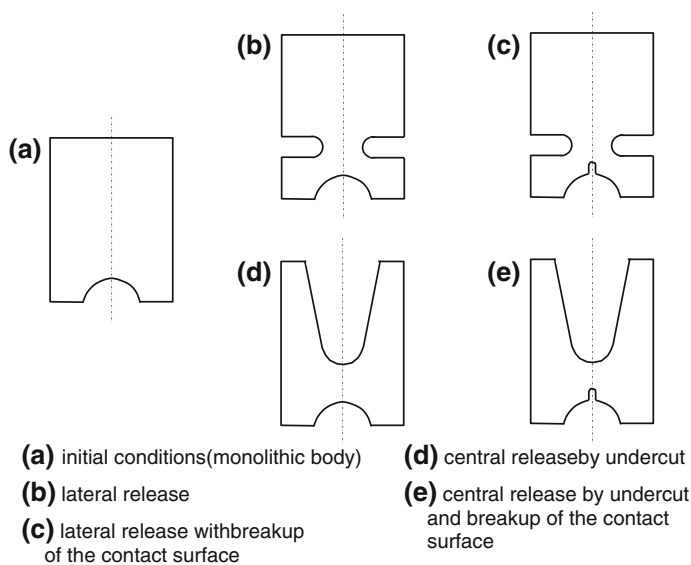


Fig. 8.18 Different release zones of tools to increase osculation

8.2.4.2 Coating and Structuring of the Tool Surface

To realize a dry manufacturing process with respect to lubrication in minimal amounts, a coating on the workpiece material could provide a reasonable contribution. This is particularly for the reduction of adhesive wear, due to the affinity between the material of the workpiece and the tools. Furthermore the friction conditions in the different zones of the tools (reduction and calibration areas) can be affected. To produce a proper surface quality, a small coefficient of friction is needed in the calibration zone. Coatings of chrome nitrite (CrN) and DLC (diamond-like

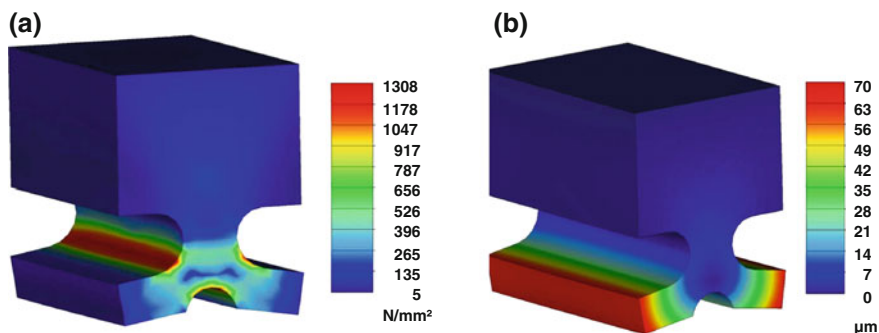


Fig. 8.19 Comparison stress σ_c **a** and overall deformation **b** of the coordinated mechanical release of a tool segment

carbon) which are already used for forming tools, e.g. with the deep drawing process, are the focus of current research. Coatings of DLC feature a coefficient of friction μ_{DLC} below 0.2, compared with CrN-coatings with μ_{CrN} of about 0.5. In particular, with DLC coatings a high resistance to wear can be expected due to its hardness, combined with sufficient ductility. Another approach to adjust the coefficient of friction in the reduction zone is the defined structuring of the tool surface by metal cutting processes, e.g. micro milling. On the other hand, in the macro area a coating can be realized by the application of Wolfram carbide (WC).

8.2.4.3 Closure of the Gaps Between the Tools for Small Component Geometries

As mentioned in Sect. 4.1, the risk that micro components can enter the gap between the dies increases with smaller diameters (Fig. 8.20). To enlarge the tool stroke, design modifications as well as process oriented methods can be considered. To avoid the undesired intrusion of the workpiece or part of it into the gap, flexible strips that offer an unaffected tool stroke, or rigid blocking elements that are guided in notches can be applied (Fig. 8.20).

8.2.5 Design of Micro Cold Forging Machines

For the production of cold forged components, transfer presses are normally used. According to the complexity of the workpieces, the forming can be designed as a single or multi-stage process. While the transport between different forming stations in the macro range can be provided by a transfer system, e.g. a gripper rail, the handling of micro parts is a major challenge which can be solved by inverting this process. An approach for a multi-stage cold forging process of micro parts is to supply the workpiece at various machining operations (stations) in the same

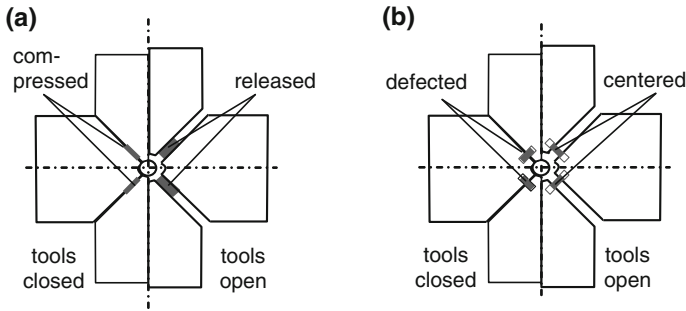
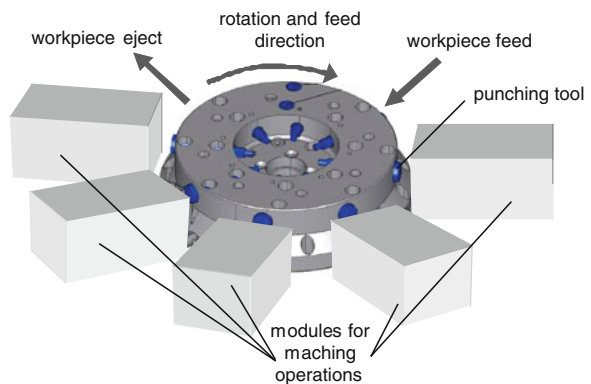


Fig. 8.20 Design oriented approaches to closing gaps between tools by **a** flexible strips, **b** rigid blocking elements

punching tool, instead of transferring it to different process stages by an additional handling system. This can be appropriately realized by arranging the punching tool on a rotary indexing table (rotor), where the different processing stations are located in the circumferential direction (Fig. 8.21).

Besides the fundamental advantage that transportation of the micro-components between the processing stations with an additional handling system is not required, the circular arrangement offers an effective material flow (components) as well as an optimized cycle time due to unidirectional rotation. Furthermore, within this machine concept different forming processes with respect to different manufacturing methods, such as forming and cutting, can be combined. The machine system shown is designed to produce small parts in a diameter range of 0.5–1.5 mm. The maximum feeding length is limited to 12 mm, and the plunger-/ejector stroke is specified at 8 mm. The press supplies a maximum force of 13 kN, and depending on the component a processing performance of up to 200 workpieces per minute can be achieved.

Fig. 8.21 Design of a forging machine for micro components (reference: Wafios, machine type SMR)



8.3 Handling of Microparts

Kirsten Tracht and Florian Weikert

8.3.1 Challenges and Principles of Micro Handling

Handling devices like grippers, storage devices and clamping devices are essential for the support of handling tasks. On the one hand the devices are component specific, and on the other they depend on the assembly station. For example, serial assembly grippers, clamping devices and feeding concepts have to satisfy several boundary conditions. In micro production, some of these restrictions and requirements are as follows:

- surface to volume ratio
- adaptability and convertibility
- clean room suitability.

Surface to volume ratio means that the smaller the dimensions of the components the smaller the ratio of the weight forces and surface forces. Adhesion of the parts to the surfaces of the gripper has to be considered during handling. Micro piece parts are not robust and have touch-sensitive surface structures. When assembling products, components of different forms and dimensions have to be handled. Thus gripper, feeding devices and clamp devices have to be adaptive and convertible to handle a greater number of micro components. Due to the fact that micro production often takes place in dust-free environments, handling devices should not be a source of contamination [13].

Handling of micro parts can be divided into two main categories: piece part handling and linked parts handling. Piece part handling is used in assembly stations for the complex micro assemblies of small- and medium-sized production lots. Linked parts handling is rather more suitable for less complex assemblies of the scale and mass production of metal formed products. A further valuation can be done regarding their applicability for transport, assembly and manufacture. The distinctions described are shown in Table 8.2.

Table 8.2 Micro handling principles

	Piece part handling		Linked parts handling
Principle	Contact handling	Non contact handling	Carrier medium
Transport	Suitable	Suitable	Suitable
Assembly	Suitable	Less suitable	Suitable with limitations
Manufacture	Suitable with limitations	Not suitable	Suitable
Examples	• Mechanical gripper • Vibration conveyor • ...	• Magnetic handling • Air cushion • ...	• Belt conveying • Friction conveying • ...

8.3.2 Piece Part Handling

8.3.2.1 Micro Gripper

Gripper systems are necessary for defined and safe handling. Micro grippers are the most developed type of handling device in the case of contact micro handling systems. Typical objects that grippers have to handle are mechanical piece parts, micro optic elements and fiber elements. Mechanical piece parts are for example gears and shaft rings with edge lengths between 50 and 1,000 μm . Micro optic elements are for example spherical lenses and mirror elements with edge lengths between 100 and 700 μm . Fiber elements are glass fibers and wires with edge lengths between 20 and 1,000 μm [14].

Micro grippers offer good access to these micro parts. Releasing micro parts is more difficult than gripping [2] because of adhesion. The functionality of grippers is well understood in the macro scale. While in the macro scale (small) parts are usually gripped with conventional principles like vacuum-sucking or mechanical gripping, in the micro range static and dynamic effects of fluids, e.g. capillary forces or temporary adhesion, can also be used for contact handling [11]. These

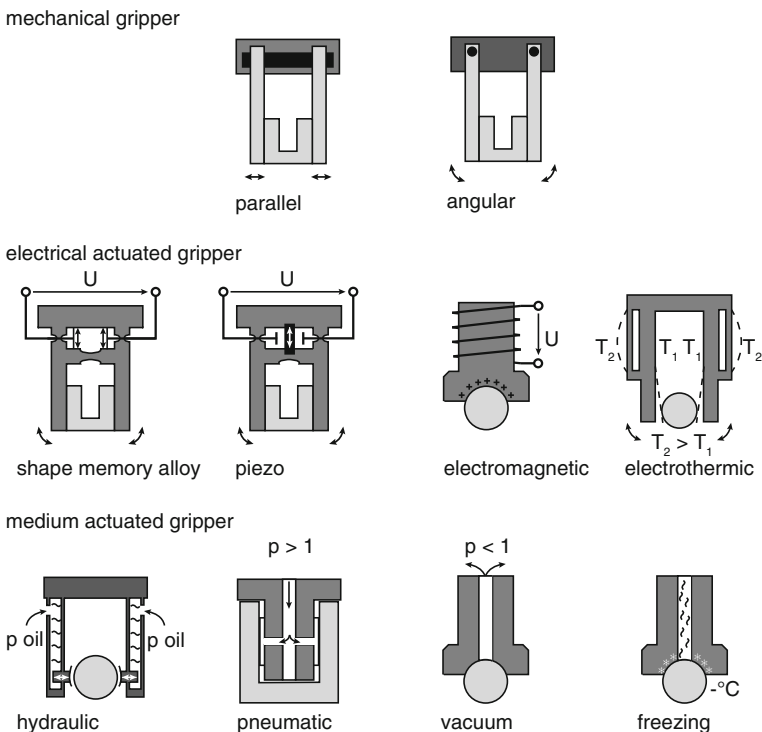


Fig. 8.22 Gripping principles

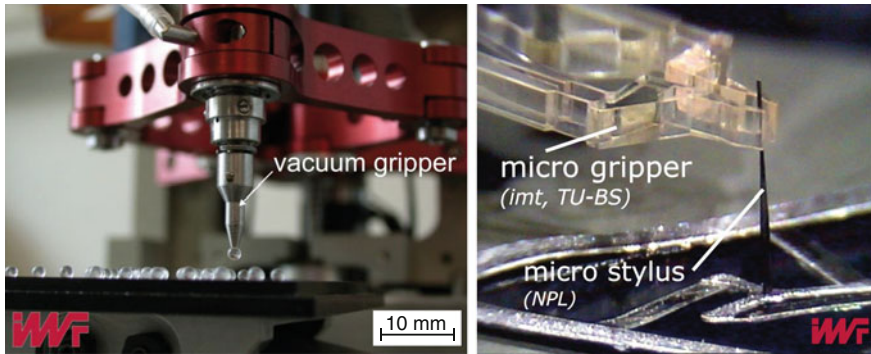


Fig. 8.23 Micro vacuum gripper (l.) [5, 8] and two jaw micro gripper [6, 7]

principles can be divided by their actuation principles. Figure 8.22 provides an overview of the main principles.

Parallel- and angular grippers are grippers where the gripping process is controlled mechanically. Grippers with shape memory alloy (sma) drives, piezo grippers, electromagnetic grippers and electrothermic grippers are electrically actuated grippers. Grippers with shape memory alloy (sma) drives use shape memory alloy wires. By applying electric potential, the gripper changes its shape and the grasping mechanism is initiated. Piezo grippers also use electric potential, which is converted into a grasping movement. Electromagnetic grippers use the magnetic field of an electromagnet. In electrothermic grippers the gripper finger is partially heated, causing bending of the gripper finger. A movement of the gripper jaw is induced but only a small movement of the fingers is possible. Medium actuated grippers use mediums like water, oil and air. Hydraulic grippers are for example operated by pressurized oil, in which the oil is within a mold padding which adapts itself to the form of the part to be handled. Pneumatic grippers use compressed air for the expansion of bellows, this induces friction between the gripper and piece part. Vacuum grippers induce a depression, within which the piece parts are sucked. A freezing gripper freezes the part to the gripper, using a freezable liquid.

Figure 8.23 shows two practical examples of micro contact grippers. On the left-hand side a micro vacuum gripper is shown which picks and places micro glass marbles. For the handling of parts smaller than $500\ \mu\text{m}$ a two jaw micro gripper is suitable. Using this gripper the handling of arbitrary formed parts is made possible [5–8].

8.3.2.2 Feeding Systems

Feeding systems are necessary for the oriented provision of the parts. Using gripping technologies requires the supply and separation of parts or small assemblies. Part supply is a major bottleneck for successful automation in micro assembly. The

feeding task can be subdivided into main parts separation, transfer, orientation and the positioning of the parts. Only a few approaches exist for adapting storage devices, order devices and transport devices to micro world conditions. At least within experimental plants, micro parts are manually sorted into face magazines like standardized waffle-packs or magazines with part adjusted molding posts. After that they are fed and positioned manually into the workspace, e.g. by a robot. An alternative to part adjusted molding posts are adhesive carrier foils. This kind of feeding is sufficient for experimental plants. For industrial approaches it is not economically feasible. Magazines, workpiece carriers and transport containers for automatic feeding, which are also cleanroom-compatible, are key issues. This includes the exact fixation of sensitive micro parts. Moreover, this requires sufficient flexibility, e.g. with regard to the form and size of the parts [16].

Different physical effects can be used for transportation. Vibration-based systems have a large share of different configurations. For the separation and supply of parts from loose bulks, vibration conveyors can be used. They are periodically induced mass-spring systems which transfer vibrations to the piece parts. The piece parts start to vibrate and hop and slide towards the intended direction. The throw principle and slide principle can be observed. The throw principle uses the combined vertical and horizontal oscillations of the feeder. The parts on the panel are induced to move forward by periodic throws. The slide principle avoids throws which is why the micro part is sliding. Both principles are shown in Fig. 8.24. By avoiding micro throws some undesirable impacts on the micro parts are also avoided. Thin and fragile parts in particular can be damaged using the throw principle, resulting in a conflict of cautious handling versus conveying speed. These problems can be minimized using the slide principle. Moreover, the slide principle offers the feeding of very different components. The parts can be transported, positioned and oriented regardless of their shape and geometry. Moreover small amplitudes of slide conveyance also facilitate a smooth conveyance [10].

In summary, vibration conveyors are widespread and are important for macro feeding systems. The drive works without sliding and rubbing parts. Therefore they are almost maintenance-free. They are easy to control and are energy efficient because of working near the eigenfrequency. Vibration conveyors are robust, have a simple design and are cheap [16]. Most existing automatic feeding systems are

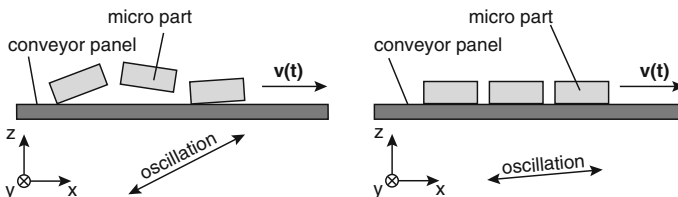


Fig. 8.24 Throw principle (l.) and slide principle (r.) [10]

designed for special applications [13]. Other principles of feeding systems in micro production include maintaining the sequence of the parts produced.

8.3.2.3 Precision Robots

Kinematics are mostly accomplished by precision robots. Precision robots are central units of micro assembly stations. There are two main embodiments that can be distinguished. The first type are robots based on serial structures, and the second type are robots based on parallel structures. Serial robots, which represent the majority of industrial robots, are based on open kinematic chains between the base and working platform (Fig. 8.25(l.)).

All joints are active and have a single degree of freedom. A large workspace is covered in proportion to the robot size. However, using serial robots also has its drawbacks. One example is the installation space of the robots. The motors of the joints also have to be moved with the robotic arm and the parts. Furthermore, the miniaturization of the robots themselves is only possible to a certain degree. Parallel robots can solve the problems of serial robots in micro assembly. Parallel robots constitute a mechanism consisting of multiple closed kinematic chains between the base frame and working platform. A closed kinematic chain has more than two links, which are closed and connected over the joints and the base frame (Fig. 8.25(r.)) [18]. A comparison of the use of serial and parallel robots in micro production is shown in Table 8.3.

8.3.2.4 Non Contact Piece Part Handling

An alternative to contact based handling systems in the micro range are systems which enable non contact piece part handling. Using magnetic and electrostatic fields are examples of handling without contact with the workpiece. Besides magnetic and electrostatic handling there are also optical levitation, Bernoulli

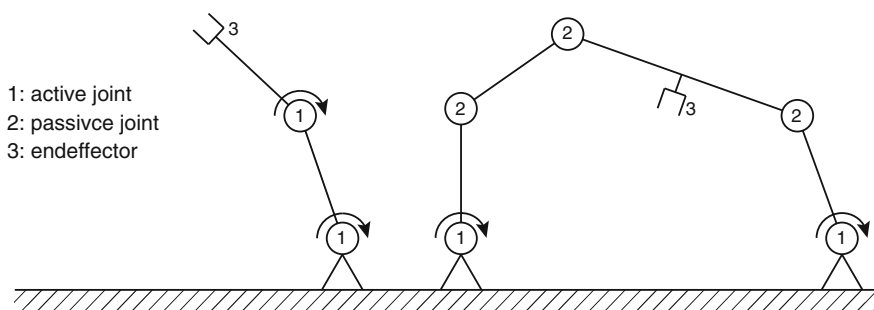


Fig. 8.25 Open kinematic chain (l.) and closed kinematic chain (r.)

Table 8.3 Suitability comparison for the adoption of serial and parallel robots in micro production [18]

Performance indicator	Serial robots	Parallel robots
Elbow room	Large (+)	Small (–)
Workspace	Large (+)	Small (–)
Installation space	Large (–)	Miniaturizable (+)
Repeatability	<1 µm possible	<1 µm possible
Clean room suitability	Yes	Yes
Velocity (v)	High	Very high
Stiffness	Low (–)	High (+)

levitation, air cushion, standing wave and near field handling. These principles are especially suitable for the feeding of micro piece parts, and are shown in Fig. 8.26.

Static electrical fields are used for electrostatic levitation, this principle enables the attraction and orientation of micro parts. The optical levitation principle uses forces of radiation pressure from visible laser light, this also enables the movement of micro parts. Bernoulli levitation uses compressed air to hold the micro part below the end effector of a manipulator which consists of a radial air outflow nozzle. The air flows downwards through the nozzle and, after having passed the micro part, flows radially outwards. Air cushioning also uses air; the air flows upwards through several holes positioned in the holding area of the gripper causing a repulsive levitation force. The repulsive levitation force counterbalances the weight of the part, which is held above the manipulator. Standing wave levitation uses an acoustic standing wave, where at the pressure nodes of the wave small components can be levitated. The standing wave is induced between a vibrating plate and a reflector. An alternative of acoustic levitation is the replacement of the reflector by the piece part itself. The main advantage of non contact handling is that surface forces can be completely neglected. Moreover, a reduction in the friction effect can provide high resolution and accuracy and prevents stick slip effects. Furthermore, a very important advantage is the absence of high local

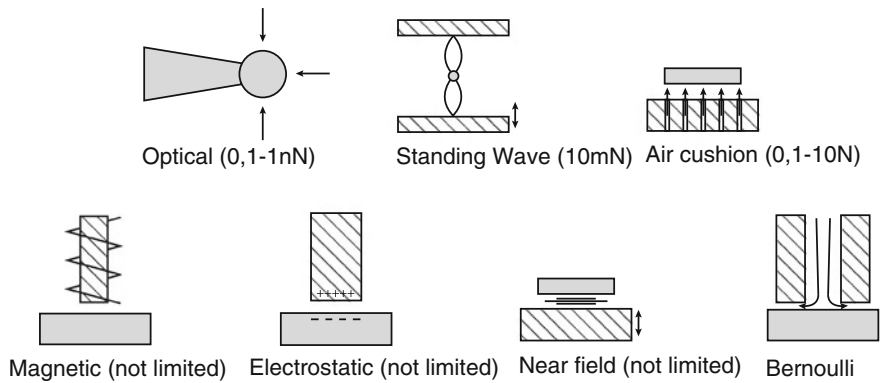


Fig. 8.26 Contactless handling principles [25]

contact pressure due to direct contact between the piece part and handling device being avoided. The risk of destroying damageable components during handling, as well as the risk of contamination from and of the end-effector, is reduced. Otherwise there are some restrictions to the applicability of non contact handling principles. (Electro)-magnetic and electrostatic principles, for example, depend on the material type. Optical levitation, air cushion and standing waves are restricted by the attainable maximum levitation forces. Optical levitation achieves a maximum of 0.1–1 nN, a standing wave 10 mN and the air cushion of 0.1–10 N. Bernoulli levitation, near field and air cushion also need planar parts or parts with at least one flat surface. Moreover, there are also environmental restrictions; (electro)-magnetic and electrostatic principles are restricted to low temperature applications, and Bernoulli, air cushion, standing wave and near field are not applicable in a vacuum [25].

Beside the restrictions mentioned, the stability of the different principles does not always maintain the required process reliability. When overcoming the actual existing challenges, non contact piece part handling can be an alternative in for example the containerless processing of materials, or within handling and transfer systems.

8.3.3 Linked Parts Handling

8.3.3.1 Basic Forms of Linkages

In mass production, linked parts are used. For the easier conveyance of linked parts the parts/blanks are kept interconnected and are not separated from the wire/strap of metal sheet/foil throughout the entire production chain. Linked parts are common when forming with progressive tools. In particular, they are very established in the production of punched parts. In micro deep drawing activities, preliminary approaches are known where the parts are transported and produced within a linkage.

The advantages of the feeding and transportation of micro parts in linkages are obvious. Due to the linkage of the parts they can be continually fed through multistage manufacturing processes when using adequate material-handling technology. Moreover, the small dimensions of the parts require the reconsideration of conventional transport strategies. Related to the small dimensions of the parts, size effects can occur which can be overcome by linking the parts. After linking they are handled collectively as a large macro part [24]. In industrial practice, linked parts are often connected to a carrier medium which enables the contact handling of the linkage without contacting the part itself. The carrier medium can be formed by a wire or a sheet or by plastic.

Many linkage designs are possible. They depend on the production process and the transported part itself. Nonetheless, three basic forms are recognizable from which nearly every real design can be derived. According to their optical appearance,

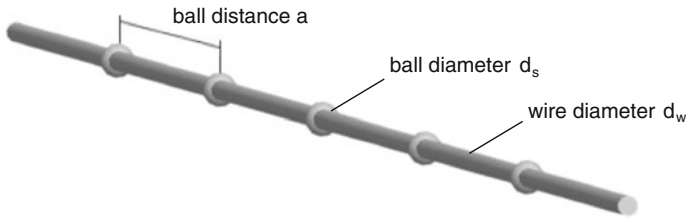


Fig. 8.27 Spheres on line linked parts

line linked parts, ladder linked parts and comb linked parts can be distinguished. Comb linked parts represent a hybrid form of the line type and the ladder type. The following remarks are confined to line linked parts and ladder linked parts, because the main characteristics of comb linked parts are similar. Line linked parts have a linear construction and are especially suitable for manufacturing products from wire [15]. The micro parts are arranged in the flux of forces. Important geometry data for processing and positioning are for example the distance between the parts, the diameter of the parts and the diameter of the wire blank with respect to the carrier medium. An example is shown in Fig. 8.27, i.e. spheres on a wire.

These spheres can for example be produced by using micro free form heading and incremental forming.

Ladder linked parts are named after their ladder-shaped construction. They are suitable for the manufacture of objects from metal strips. The parts are arranged in the center of the strip. The fringes of the strip are used for conveying and the connection of the linkage. Transport forces are applied to the fringes of the strips. The fringes of the strips are used for transport. The distances between the transport areas and forming zone must be large enough to prevent the mutual impacts of forming and conveying [15]. The principle of ladder linked parts is shown in Fig. 8.28. Within this example micro cups are manufactured within metal strips using deep drawing/stretch forming and punching.

A special challenge is the conveyance and feeding of the linkages. Therefore the conveying zone, which is not necessary for forming the product, can be used. This means that additional material is now used as the transport string. Often support materials like plastic belts are used as carrier medium. In micro sheet shaping the fringes of the sheet are already often used for conveying. In both cases the belts are fed into progressive tools or micro assembly stations. Progressive tools are tools where cutting operations using the principle of progressive forming procedures, like bending, coining or upsetting, are accomplished in a second or third step.

In the production of electronic and fine mechanic components or complete products in large lot sizes, the processes for handling and part inspection can be integrated. Examples of these processes are order devices, handling and feed systems, assembly operations and cleaning operations.

Such production systems have some special advantages that overcome problems of the handling of single piece micro parts. Single piece parts are often

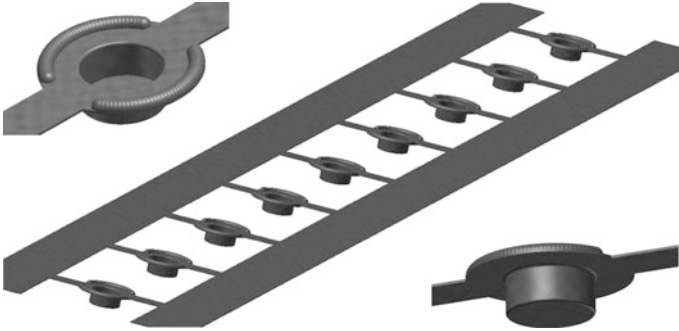


Fig. 8.28 Ladder linked parts

difficult to handle in ordering and positioning procedures in automatic assembly. Cycle times become longer and it is difficult to cope with disturbances. Using progressive production systems these difficulties are avoided, because multiple production procedures are handled in one system. Due to the keeping of parts in a linkage, bulk material is avoided. The higher the integration degree of part production, the shorter the cycle time becomes. Industrial companies already handle linked parts successfully. However, part dimensions of less than 1 mm in at least two dimensions cannot be handled reliably. Figure 8.29 shows an example of a progressive tool and linked parts.

Not only does the production system need to be designed, but also the parts have to be designed for linkage production. The selection of the right material is of great importance. Moreover economic restrictions like a lot size of more than 2.5 million parts a year have to be taken into account.

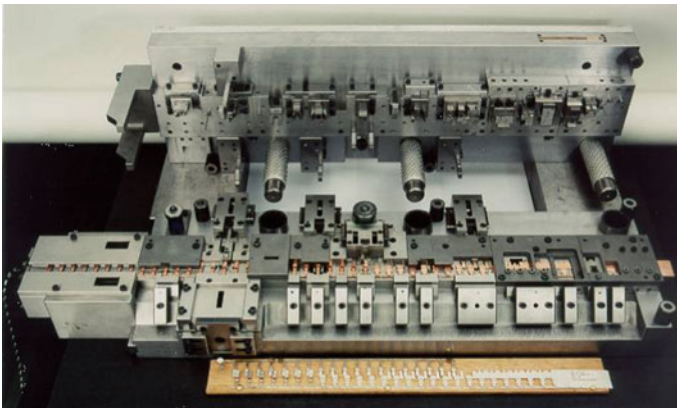


Fig. 8.29 Progressive tool and micro parts linkage [tyco electronics]

8.3.3.2 Synchronization of Linked Parts

The adjustment and synchronization of multistage manufacturing systems is a great challenge in micro production. By using different technologies, different set-up times have to be coordinated. Moreover ramp up losses have to be minimized and process reliabilities have to be maximized.

When manufacturing the linkage, structure procedural requirements and restrictions like accessibility, holding forces and clamp forces, the forming zone itself, axial material flow and many more have to be taken into account. It is important to synchronize good parts using linkages, for example within assembly stations. Hence, a big challenge during manufacturing and transportation is the synchronization of the different processes [23]. Compared to one-stage process, the planning and coordination efforts are higher. Examples are the planning of the lot size and assembly sequence. Poor planning and coordination worsen the processing time. Since the development of analytic models is a key function in the coordination of material flow, it has a great practical value [3].

Nonetheless, the advantages of linking parts outweigh the disadvantages, as well as the planning and coordination efforts. These efforts are profitable, especially for products that have to be produced in large lot sizes with the involvement of several manufacturing processes.

Technical solutions for the synchronization of movements already exist. There are controlled and regulated systems to mechanically balance assembly objects and convey parts. Mechanical synchronization of conveying systems, handling systems and objects are realized by form fit connections or by force fit connections. Controlled synchronization is achievable by specifying the nominal values for the conveying direction and handling direction, whereas regulated synchronization accounts for additional relative derivations. It is very important to adhere to the synchronization distance, integration methods, synchronization principles and synchronization run [26].

8.3.3.3 Strategies for the Extraction of Scrap Parts from Linked Parts

Defect parts in micro production become a problem, because of their potential to jeopardize process reliability. They can stick to the tools because of van der Waals forces and damage other tools and parts. In precision mechanical engineering, where an exact orientation and a precise position are vital, defects are time and cost critical.

A possibility to avoid these problems and to only supply good parts is to separate the defects from the linkage. But ladder linked parts and line linked parts can lose connection by the use of cutting operations. Connection has to be restored by adequate joining measures. A great challenge is to cut and rejoin without deformation, because tolerances are of only a few μm . In addition, maintaining of the position during cutting and assembly processes and repositioning after cutting are other important factors to be taken into account. Positional changes cause imprecision in the subsequent process, and in the worst case more defects.

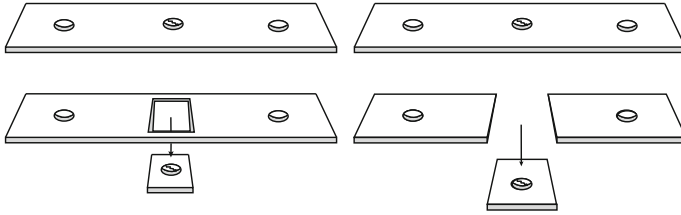


Fig. 8.30 Cutting principles without losing connection (l.) with loss of connection (r.)

The first option for removing scrap parts from linkages is to separate defective parts without breaking the connection of the linkages. This option is shown in Fig. 8.30. Tolerances of a product are affected by the accuracy of positioning in handling. Exact positioning, however, is very difficult because of the small dimensions of the parts. Otherwise it is convenient that the distances between the parts and their orientation remain unaltered. Another disadvantage is that a gap emerges at the linkage. Gaps are weak spots. In the case of cracking of the linkage because of material failure, gaps can cause standstills within the process chain. Even more problematic is that at the following process step a part may be missing for an assembly operation. This has to be considered when programming the machine control.

A second possibility is to shorten the linkage (Fig. 8.30(r.)). The predefined distances between the parts can be ensured by separating broken parts, including their carrier medium, from the linkage. Thus the complexity of control engineering, data handling and sensor systems of the following process steps are reduced. The number of process steps for the separation and rejoining of the linkage at one processing station is challenging.

Two alternatives for rejoining linkages are shown in Fig. 8.31. Either the two ends of the linkage are positioned edge-to-edge and are rejoined by a welding operation, or they are joined overlapping by a forming operation. Cutting and rejoining of the linkages is a durable solution to deal with defects. Efforts in handling during the following process steps, especially in long process chains, are reduced.

8.3.4 Accuracies and Tolerances in Micro Handling

Micro production is confronted with high process variability and an increased significance of measurement uncertainty in relation to tight tolerance specifications [9]. The consideration of these tolerances and accuracies in micro production is an element of holistic balancing of part features and process characteristics. They play an important role in micro handling. In micro production and handling a trend towards rising accuracy is recognizable, depending on the application. In recent years the range lies between 0.01 and 20 μm [1].

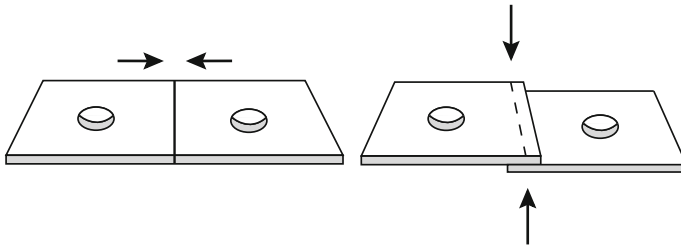


Fig. 8.31 Joining principles without (l.) and with (r.) overlapping

Using precision robots as an example, the handling accuracy has to be very precise. But, the accuracy of mechanical precision robots is limited. Often disturbing influences like fabrication defects, friction, thermal expansion or control errors play a greater role in disturbance than in the macro world. Regular maintenance and calibration activities need greater efforts than in the macro scale, which makes them more expensive [4].

Nonetheless, gripping operations achieve an adequate accuracy when using self centering principles. Also sensor systems for readjustments are realizable. In cases where the functionality of a micro product depends on the correct distance and the correct orientation of parts of an assembly group, the accuracy of the distances is very important. An example is the retention of distance and parallelism between two slides. DIN 7157 defines tolerance zones according to the dimensions of the part. Tolerance zones also become smaller when the part dimensions become smaller. Thus, challenges in manufacturing precision increase when part sizes shrink. That means that part dimensions in the millimeter range use or need tolerances in the micro range. But, defining tolerances for micro products is restrained by physical limits. Tolerances cannot be reduced by the same percentage as part dimensions. Furthermore, particularly at fits the surface roughness has an increasing impact [18].

These examples show that the standards for the macro range cannot be used unrestricted in the micro range. Hence, presently tolerances in micro production planning, and thus for micro handling, have to be defined for every individual case.

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Chapter 9

Simulation Technologies

Peter Maaß

Latin

a_{pc}	Depth of cut (μm)
a_{pc}^1, a_{pc}^2	Given depth of cut (μm)
$a_{pc}^{\text{act}}(t)$	Real depth of cut (μm)
a_{pc}^{var}	Variation of the depth of cut (μm)
$a_{pc\alpha}$	Optimal depth of cut (μm)
$a_{pc\alpha}^1$	Optimal depth of cut for a_{pc}^1 (μm)
$a_{pc\alpha}^2$	Optimal depth of cut for a_{pc}^2 (μm)
Δa_{pc}	Difference between the given a_{pc}^1 und a_{pc}^2 , the height of the ramp (μm)
a, b	Parameters that describe the stochastic spread of the grain orientations around a basic orientation
B	Force coefficient
C	Stiffness matrix (MPa)
C	Stiffness matrix of polycrystal (GPa)
c	Stiffness matrix of monocrystal (GPa)
F	Force (N)
F_o	Surface force (N)
F_t, F_r, F_{ax}	Force in tangential, radial and axial direction (N)
F_x, F_y, F_z	Resulting force in three orthogonal directions x, y and z (N)
F^*	Vector of three force components (N)
f	Body forces (N/m^3)
$g, g_o, \tilde{g}$	Rotation

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h_c	Chip thickness (mm)
i	Refraction index
j	Number of cutting edges
k	Number of small independent rotations
l_h	Height of the holder (mm)
L_q	Lebesgue saces
m	Function
n	Rotational speed of the tool [Hz (1/s)]
q	Parameters
t	Time (s)
v	Velocity (mm/s)
v_f	Feed velocity (mm/min)
v_f^{act}	Current feed velocity (mm/s)
w	Acement/deformation vector (m)
x	Coordinate in x-direction (mm)
y	Coordinate in y-direction (mm)
z	Coordinate in z-direction (mm)

Greek

A	Operator
α	Regularisation parameter
γ	run out angle [$^{\circ}$ (deg)]
δ	Deflection (μm)
$\delta_t, \delta_r, \delta_{ax}$	Deflection in radial, tangential and axial direction (μm)
δ_v	Virtual displacement (m)
Δ	Difference (μm)
Δx	Length of the ramp (mm)
ε	strain tensor
$\varepsilon, \varepsilon_{el}, \varepsilon_{in}$	Strain tensors
$\varepsilon_{xx}, \varepsilon_{yy}, \varepsilon_{zz}, \varepsilon_{xy}, \varepsilon_{xz}, \varepsilon_{yz}$	strain tensor components
ζ	orthogonal basis
$\theta_1, \theta_2, \theta_3$	Euler angle (rad)
η	run out vector (μm)
η_x	Run out vector in x direction (μm)
η_y	Run out vector in y direction (μm)
v	variable
ξ	Integration variable
ρ	Density (kg/m^3)
σ	Stress tensor (MPa)
σ_{ij}	components of stress tensor (MPa)
$\bar{\sigma}$	Averaged Stress tensor in RVE (MPa)

Φ	Penalty functional
ϕ	accumulated plastic strain
φ	Rotational angle [$^{\circ}$ (Deg)]
Ω	Rotation matrix

9.1 FEM-Simulation

Jonathan Montalvo-Urquizo and Alfred Schmidt

9.1.1 Mechanical Models for Micro Metal Forming

Based on the physical principle of the conservation of momentum, the linear elastic deformation of a solid body is described by the time dependent or stationary elasticity equation

$$\rho u_{tt} - \operatorname{div}(\sigma) = f \quad \text{or} \quad -\operatorname{div}(\sigma) = f \quad (9.1)$$

in the body's three-dimensional domain Ω , subject to (initial and) boundary conditions. Here, u depicts the deformation vector, f the body forces, and $\sigma = (\sigma_{ij})$, $i, j \in \{1, 2, 3\}$ the stress tensor. For stationary situations, the first term on the left hand side vanishes.

Material laws describe the relationship between deformations and stresses. In the linear elastic model, the stress tensor can be described using Hooke's law as a linear relation between the stiffness and strain tensors. Denoting the strain tensor as $\varepsilon = (\varepsilon_{kl})$, $k, l \in \{1, 2, 3\}$, this relationship can be expressed by the tensor equation:

$$\sigma_{ij} = C_{ijkl} \varepsilon_{kl} := \sum_{k,l=1}^3 C_{ijkl} \varepsilon_{kl} \quad (9.2)$$

where $C = (C_{ijkl})$, $i, j, k, l \in \{1, 2, 3\}$ is the elasticity or stiffness tensor. Due to symmetry, the strain and stress tensors only have six independent entries which can be denoted as $\sigma_{11}, \sigma_{22}, \sigma_{33}, \sigma_{23}, \sigma_{13}, \sigma_{12}$, using the indices $\{1, 2, 3\}$, to refer to the X , Y and Z coordinates. Using these components, a matrix–vector notation can be used instead of the tensor notation. The expression above can now be written:

$$\sigma = C \varepsilon \quad (9.3)$$

where σ and ε are now vectors in R^6 and C is the stiffness matrix in $R^{6 \times 6}$. Again, the matrix C is symmetric and only has 21 independent entries.

For special materials, the stiffness matrix C only has 12 non-zero entries which can be computed from only nine, three, or even two independent elasticity constants. These special matrix forms correspond to the cases of orthotropic, cubic,

and isotropic materials, respectively. For isotropic materials, the stiffness can be computed from the modulus of elasticity E and the lateral contraction ratio ν (or, in other notation, from the two Lamé constants). In macroscopic situations, the elastic behavior of a metal workpiece can typically be well described by an isotropic material law, and the corresponding elasticity constants can be determined by simple experimental setups.

When one looks very closely, as in the case of micro components, the local mechanical behavior of a piece of metal is typically not *isotropic*. This is due to the fact that solid metal atoms are arranged in a crystal lattice of e.g., cubic type, and the elastic behavior in the lattice directions is different to behavior in non-lattice directions. Thus the *local* elastic behavior is typically *anisotropic*, and this anisotropy is, except for a very small and negligible deviation, constant inside each metal grain.

Due to the fact that there are many different grains in a macroscopic workpiece, with anisotropies in many different and randomly distributed directions, these anisotropies average out and the macroscopically observable behavior of the metal is isotropic. Thus, in the macro scale setting, the effects of the material microstructure can typically be neglected, and homogenized, isotropic material behavior is a good approximation to describe the macroscopic mechanical response of a workpiece. In special situations (like a material with segregations, etc.), the assumption of an isotropic response is not enough to describe the macroscopic behavior with sufficient accuracy, and an anisotropic model must be used even in the macroscopic scale.

But for micro components, the dimension of the component and the size of grains might be comparable, and there might be only very few grains, especially in the transversal direction of a metal foil. An assumption that the local anisotropies do average out is no longer valid, and the effect of single grains is observable, with deformations and stresses varying from grain to grain. Thus, the two Lamé constants are not enough to determine the appropriate stiffness matrix C and a more general material model with typically 6 elasticity constants must be used. The three elasticity constants for a cubic monocrystal in lattice orientation can be determined by experiments and give an elasticity matrix c . For a grain with a general orientation, the stiffness matrix C is then given by the multiplication of the corresponding rotation matrix g with the matrix c as:

$$C = gc. \quad (9.4)$$

A general three-dimensional rotation can be described by three angles. For this reason the total number of degrees of freedom for C is six for the case of a cubic crystal material. For a processed piece of material, like a rolled sheet, typically some orientations are more probable than others. These preferences in orientation are described by the texture of the material (compare with Sect. 9.2.1).

Forming processes produce *permanent deformations* of the workpiece, also called *plastic* deformations in material science literature. In a continuum mechanics model, plastic deformations occur when the elastic stress reaches

values larger than the so-called elastic limit or yield stress σ_y . In formulas, the yield stress is given as an upper bound for the *equivalent stress*, which is the norm of the stress deviator defined as

$$\sigma^* = \sigma - \frac{1}{3}\text{trace}(\sigma)I, \quad (9.5)$$

with I denoting the identity matrix in three dimensions. Using this definition of equivalent stress gives the well known *flow rule*

$$\|\sigma^*\| \leq \sigma_y, \quad (9.6)$$

meaning that any stress tensor with a norm over the yield value cannot be a valid stress. In this case one speaks about material flowing. Material is said to flow as it undergoes plastic deformation instead of increasing its internal stored stress. In other words, the creation of plastic strains in the material compensates the stress level during strong loading.

The yield stress can vary during a plastic deformation process in the presence of hardening. For isotropic materials, this norm is typically just the von Mises norm. This is why the equivalent stress is also known as von Mises stress. But for anisotropic materials, the norm can also be considered anisotropic.

At molecular scale, permanent deformations are generated by the dislocations of planes in the crystal lattice. When the deformation model respects the orientation of the local crystal lattice, then one may even try to model possible dislocations along various planes and compute plastic deformations in such a manner.

9.1.2 Mathematical Foundations and Numerical Implementations

Different approaches exist for the numerical approximation of the system of partial differential equations (PDE)—finite difference methods, based on the approximation of derivatives by difference quotients on a structured grid, and finite element methods, based on a (typically piecewise polynomial) approximation of functions on unstructured meshes (for example using tetrahedral, hexahedral, or mixed geometric cells). While the first one can exploit the grid structure for an easier, efficient implementation and solution, the latter one offers much more flexibility in choosing grids, especially for non-rectangular domains and specialized function spaces, already incorporating the desired local properties of the problem solution. Here we will restrict ourselves to the description and presentation of some finite element methods.

While from a mathematical point of view the derivation of finite element formulations is based on a weak, variational formulation of the system of partial differential equations, a mechanical interpretation of this is given by the virtual work. The virtual work setting considers the problem of finding the amount of

work produced by some virtual deformations, which are small variations of the body's deformation restricted by e.g. the boundary conditions of the mechanical system.

For the three-dimensional case of a body at static mechanical equilibrium, one considers an arbitrary vector-valued displacement function, δv , and calculates the amount of work in the body. This is the sum of the forces acting in the body times the corresponding displacement, it is:

$$\int_{\Omega} (\operatorname{div} \sigma + f) \delta v dV = 0. \quad (9.7)$$

with the equation being equal to zero as the expression in parenthesis is zero. This equation can be transformed using the divergence theorem into:

$$\int_{\partial\Omega} \sigma \cdot \delta v ds - \int_{\Omega} \sigma \cdot \operatorname{grad} \delta v dV + \int_{\Omega} f \delta v dV = 0, \quad (9.8)$$

where $\partial\Omega$ represents the boundary surface and σ the traction forces on it. Recalling the arbitrariness of the virtual displacement δv , we can think of a displacement increment as a measure of a virtual strain $\delta \varepsilon = (\nabla \delta v)_{\operatorname{sym}}$. The equation of virtual work becomes

$$\int_{\partial\Omega} \sigma \cdot \delta v dS - \int_{\Omega} \sigma \cdot \delta \varepsilon dV + \int_{\Omega} f \delta v dV = 0. \quad (9.9)$$

This last equation is also known as weak formulation in the context of mathematical variational methods for which the virtual displacement vector is taken as a general function belonging to a certain function space, usually of polynomial type. The numerical solution of the virtual work equation (or the weak formulation) is done by representing the virtual displacement in terms of a basis of the function space containing it and using this representation to split each term in the equation as a sum. These sums can be later ordered to form a linear system of the form

$$Au = F, \quad (9.10)$$

where the elements of the matrix A include the information from the first two terms in Eq. (9.9), and the vector F includes the terms from the last integral.

This solution approach is known as Galerkin method to solve the weak formulation of mechanical equilibrium, and constitutes the theoretical support for the Finite Element Method (FEM). In practical applications, FEM takes special forms of the basis functions with local support, so that many of the integrals over products of basis functions, which are needed to build the linear system, are zero. This approach results in sparse and well structured matrices for the linear system in Eq. (9.10) and has major advantages for the efficiency of the method.

For the FEM solution of mechanical equilibrium, the expression of stress σ in terms of strains is required. For problems in which the strain levels are very small,

the generalized Hooke's law for anisotropic elastic behavior fully describes the relation between strain and stress, but for larger forces and/or deformations this relation must be described using a decomposition of the strain in an elastic (ε_{el}) and an inelastic part (ε_{in}). The most common approaches are to decompose the strain additively as:

$$\varepsilon = \varepsilon_{el} + \varepsilon_{in}, \quad (9.11)$$

or multiplicatively as:

$$\varepsilon = \varepsilon_{el} \varepsilon_{in}. \quad (9.12)$$

This gives a different form of the weak formulation of the problem, as the equilibrium equation holds for the elastic part of the strain. Both approaches are widely used, the additive for problems with small deformation gradients, and the multiplicative for problems with large deformations, as in the case of forming.

Numerically, the elasto-plastic problem is solved by means of a method of predictor–corrector type as described in mathematical terms by [1]. A more practical description of the method's implementation can be found in [2]. The predictor–corrector method is applied for a time dependent mechanical problem in which some forces are applied at every time step. It assumes that all variables of the mechanical problem are known up to the previous time step, and computes their evolution in the current time step. The evolution of all variables is updated in one time step by the following procedure for every single node in the numerical grid:

1. **Initialization** Assume that the variables of elastic strain, inelastic strain, and stress are known for all points in the numerical domain for the simulation, i.e., $\varepsilon_{el}^n, \varepsilon_{in}^n, \sigma^n$. Also assume that the values of accumulated plastic strain ϕ^n from all previous time steps have been stored.
2. **Elastic predictor** The loads of the current time step are applied and the stress in the current time step is computed using the approximation of the elastic stress. This produces a prediction for the correct stress assuming that the loads in the material remain in the elastic range.
3. **Flow rule** The predicted value is evaluated with the flow rule to decide whether the value is under or above the yield stress value. This labels the current node (at the current time step) as elastic or plastic, and decides between one of the following two stages in the algorithm.
4. **Elastic point** If the stress prediction is lower or equal than the yield stress, the time step is declared elastic for the current grid node. Predicted values are then the correct values and the variables for the current time step can be updated as

$$\sigma^{n+1} = \sigma_{trial}^{n+1}, \quad (9.13)$$

$$\phi^{n+1} = \phi^n. \quad (9.14)$$

After the elastic update has been performed for the strain and stress variables, the calculations at this point are finished.

5. **Plastic point** In the contrasting case that the prediction is above the yield stress, the predicted values must be corrected as any stress above the yield value is not allowed. This is done by updating it by means of e.g., a Newton–Raphson scheme [3], giving the values for σ^{n+1} and ϕ^{n+1} which are consistent with the flow rule. The updating procedure at the node finishes when the consistent values have been found. After that, the new values of the variables are stored and the calculation in this point is finished.

This point wise update procedure substantially increases the computational work required when compared to pure elastic simulations. The algorithm is valid for both the additive and the multiplicative decomposed forms of strain, cf. Eqs. (9.11) and (9.12). The resulting forms of the equations for both cases, as well as other types of inelastic material behavior like hyperelasticity or viscoplasticity, can be found in [4] in a rather mathematical setting, or in [2] with an emphasis on ideas for computational implementation.

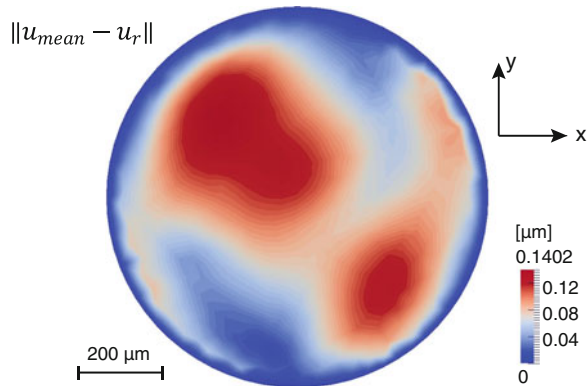
9.1.3 The Role of Different Scales in the Simulation of Micro Components

When considering metallic micro components, the effects of local differences in the material properties cannot be neglected. For example, it is usually assumed that an averaged elastic modulus exists for every metallic material. This idea is based in the existence of millions of material grains whose properties are smoothed out to give the averaged modulus.

In the case of micro components, their sizes do not allow the existence of such a large amount of grains. Differences in the local material properties cause unavoidable differences in material behavior during the production and use of micro components. In this sense, every simulation strategy using averaged material properties will only give a very coarse approximation to the possible behavior of any micro component.

In order to illustrate this idea, the simulation of deformation on a piece of steel foil (type DC01, thickness 50 μm) has been performed for a test case of pressure under the foil. Figure 9.1 shows the discrepancy between the deformation obtained via a simulation with the stiffness averaged value and another simulation using realistic assumptions on the local properties. It can be observed how the slight stiffness variations cause discrepancies of up to 0.14 μm all over the foil piece. The local stiffness has been chosen using the texture properties of the DC01 foil (cf. Sect. 9.2). This result illustrates the necessity of considering local differences in the material properties, which can produce relatively large differences in the material responses at the micro components' scale.

Fig. 9.1 Euclidian norm of the deformation's difference computed using averaged stiffness' values (u_{mean}) and computed using a stochastic realization of randomly-chosen stiffness' variations (u_r)



9.1.4 Multi-Scale Finite Element Methods

The special features of micro components necessitate a simulation strategy which not only includes the component scale but also the scale of its single crystals. A common approach to simulate problems in multiple scales is the so-called FE^n method, based on the theory of mathematical homogenization, and its corresponding two-scale version FE^2 as presented in e.g., [5].

The FE^2 method solves problems in which information on two different scales must be included. The first scale corresponds to the component under consideration and the second corresponds to the scale where information about the single crystals is used. The main idea is to consider a FEM-simulation in the first scale in which the material properties at each integration point are obtained via the computation of an embedded FEM-simulation in the second scale. In computational mechanics literature the first and second scales are referred to as “macro” and “micro” scales, respectively. These names are not used here to avoid confusion in the context of micro components.

In practice, the first and second scales correspond to the component and the so-called representative volume element (RVE), respectively. The RVE consists of a volume of material small enough to avoid any smoothing of the local properties, and large enough to include at least several single material crystals. It is important for the method that the component geometry in the first scale can be described as periodic repetition of many RVEs [6]. Thus geometrical periodicity must be considered while constructing any descriptor of the RVE domain.

The FE^2 method has been successfully applied to problems of materials in two-phases, foams or materials with pores. In these applications the RVE is often constructed as a small volume made of the base material with an inclusion of the second material in the center. The proportional volume fractions or the amount of pores per unit volume are used to determine the precise size of the inclusion in the base material.

The coupling between the first and the second scale is performed via so-called localization and homogenization steps:

- During localization, for every integration point in the first scale, the variables are passed to the second scale as boundary conditions to the FEM-simulation. The solution of the FEM-simulation in the second scale produces the local material response, with all detailed features like grain geometries and lattice orientations. The results of this simulation fully describe the variables in the continuous domain of the second scale.

In the case of mechanical problems, it is common to take values for strain at a point of the first scale and to pass them to the second scale as deformations. Furthermore, these deformations are applied to the outer boundary of a small volume around the point under consideration, i.e., the RVE at this point.

- Once the variable values in the second scale have been obtained, they can be homogenized through an integration process over the RVE geometry and passed as local responses for the current integration point in the first scale. With this step, the responses of the second scale can be transformed in punctual values for the simulation of the first scale.

The method is named FE^2 due to the fact that for every integration point in the first scale, a complete simulation in the second scale must be performed. Although FE^2 methods are in general well suited to solve problems with different scales, they have the drawback of being computationally expensive. The large amount of FEM-simulations required to get one simulation in the first scale is often calculated with tailor-made parallel implementations, making neither of these solutions portable nor easy to use. An alternative to these expensive computations is the material characterization for the second scale via a large set of RVE configurations.

9.1.5 Computations in the Second Scale

For mechanical problems in micro components, calculations at the second scale (i.e., the crystal aggregate with only few material grains) need the solution of the localization problem in the RVE geometry.

The amount of grains fitting inside a RVE for metallic materials can be obtained from experimental measurements of the grain size and, if available, from certain approximations of grain size distributions [6]. For example, one could consider larger grains close to the upper and lower boundaries by assuming a distribution of grain seeds with higher density in the vertical middle plane. For a given number and distribution of single grains, the location of their centers can be stochastically generated and the grain shapes can be constructed in form of a Voronoi tessellation of the RVE [6].

In order to fulfill the strong homogenization assumption of periodicity in the RVE, the creation of randomly located grains must be implemented with this consideration. This allows the creation of RVE geometries of cubic shape in which the grains at any boundary are not only cut, but the sliced part appears in the opposite boundary, giving to any aggregate a perfect periodic structure.

The periodicity is implemented by placing ghost grains outside the RVE in periodic locations to each generated grain center (for X and Y periodic RVEs each grain has eight copies whose centers lie outside the RVE). Figure 9.2 shows two examples of RVE geometries and Fig. 9.3 shows some grains belonging to the first RVE. These geometries have been constructed for the DC01 foil with a thickness of 50 μm and include the periodicities of a foil in the X and Y directions, while the vertical direction is not periodic.

Local differences among the single crystals are determined by their stiffness values. This causes the different regions in the RVE to have a stronger/weaker reaction when mechanical loads are applied. The stiffness at each crystal can be determined using its lattice orientations and thus can be easily connected to concepts at the grain scale like texture (cf. Sect. 9.2).

Once the geometries and grain orientations are given for a RVE, all possible local material responses can be examined. For this, each single crystal in the RVE is endowed with its own mechanical properties. In this form, the stiffness through the material is constant in every crystal, but differs among the different crystals. The space-dependent stiffness matrix of the complete RVE-geometry can then be described as:

$$C(x) = C(j), \quad (9.15)$$

for x inside the j th crystal. j represents the crystal index this relation can be used for any point $x = (X, Y, Z)$ inside the domain.

The local modal responses of the material can be determined in the elastic case by deforming the geometry of the RVE and analyzing the stress results in the interior [7]. It is well known that the grain's lattices can strongly influence the stress level, which depends on whether or not they are aligned with the direction of the loads.

Figure 9.4 shows the stress responses obtained for the second RVE example in Fig. 9.2 for the cases in which the aggregate has an elongation in the X direction and shear in the XZ plane. These responses are two of the six possible aggregate

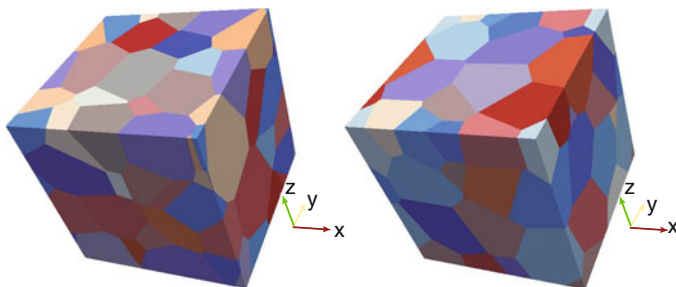


Fig. 9.2 Examples of representative volume elements. The geometries are constructed with periodic boundaries in the X and Y directions. The RVE geometries are constructed as *cubes*, with a diameter 50 μm

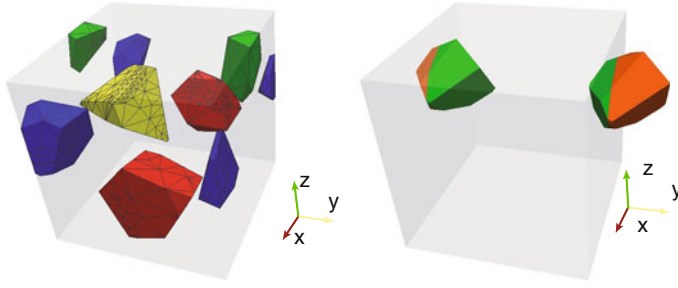


Fig. 9.3 Grains inside a RVE geometry with cuts due to periodicity. *Left* examples of grains with periodical cut in the X direction (*red*), with periodical cut in the Y direction (*green*), and with periodical cut in the X and Y directions (*blue*). Also shown is an internal grain without cuts (*yellow*). *Right* Periodicity idea, example of a grain with Y periodicity

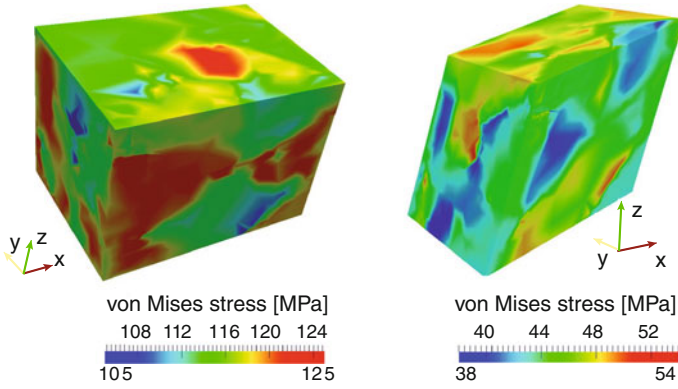


Fig. 9.4 Stress results for tension in the X direction (*left*) and shear in the XY plane (*right*) of the same aggregate. The deformation of 1 % in the aggregate is scaled by a factor of 40 to improve visualization

responses for the RVE and are obtained by solving the boundary value problem for elasticity.

$$-\operatorname{div}(\sigma(x)) = 0, \quad \text{in } \Omega \quad (9.16)$$

$$u(x) = Mx, \quad \text{on } \partial\Omega \quad (9.17)$$

with x and u being the 3D coordinates and the displacement vector, and with the stress at point x defined in the anisotropic form

$$\sigma(x) = C(x)\varepsilon(x), \quad (9.18)$$

with the stiffness matrix defined as in Eq. (9.15).

Note that the boundary condition on $\partial\Omega$ assigns a fixed outer shape to the RVE and this causes different responses (or levels of stress) in the RVE's grains. The

symmetric matrix M determines the specific form of the Dirichlet boundary condition on the aggregate geometry and is, for the cases of tension in the X direction and shear in the XZ plane, equal to

$$M_{xx} = \begin{pmatrix} \xi & 0 & 0 \\ 0 & 0 & 0 \\ 0 & 0 & 0 \end{pmatrix}, \text{ and } M_{xz} = \begin{pmatrix} 0 & 0 & \xi \\ 0 & 0 & 0 \\ \xi & 0 & 0 \end{pmatrix}, \quad (9.19)$$

with ξ denoting the amount of strain applied.

Due to symmetry, there are only six possible boundary conditions or modes to force the RVE, each of them determined by the corresponding matrix $M_{ij} \equiv M_{ji}$ for $i, j \in \{x, y, z\}$. Solving the boundary problem, Eqs. (9.16–9.17), for each of these six modes provides information on the response of the aggregate for all possible loads. Furthermore, any elastic load in the boundary of the RVE can be decomposed in a weighted sum of modal responses [7]. This idea allows the direct connection with the two-scale methods as it provides a more efficient way of computing the solution of the micro problem after localization.

The colors in Fig. 9.4 shows the resulting stresses from tension and shear applied to the second RVE in Fig. 9.2, with grain stiffness matrices taken from a texture model (cf. Sect. 9.2.1). It can be observed how the grain geometry in combination with the local stiffness determines how low/high the grain is stressed by applying different forms of load on the aggregate. Here, both strains applied to deform the boundary used $\xi = 0.01$, i.e., 1 % of deformation.

Following the ideas of an FE² method, the next step after the RVE solution (second scale) has been computed is homogenization. For this, the local material responses are concentrated and passed towards the component simulation (first scale). This is done for the loading modes on the aggregate by calculating the six averaged stress tensors, each of them having the form:

$$\bar{\sigma}_{ij} = \frac{1}{|RVE|} \int_{RVE} \sigma_{ij}(x) dx, \quad i, j \in \{1, 2, 3\}. \quad (9.20)$$

Based on the six averaged tensors, the overall response of the RVE can be described in a compact form, including all local variations in both geometric features and stiffness values of every single crystal. The components of the complete stress tensor for the second RVE in Fig. 9.2 are presented Table 9.1.

9.1.6 Crystal Plasticity Methods

In recent years there has been a strong impulse to include the plastic effects occurring in the material crystals. One of the most successful approaches to this idea is the so called Crystal Plasticity Finite Element Method [3].

The method is also based on a mathematical homogenization idea, with information on the plastic behavior of the single-crystals or crystal aggregates. The

Table 9.1 Example of averaged stress values for one RVE under tension and shear using $\xi = 0.01$

Load	Stress averaged response (MPa)					
	$\bar{\sigma}_{11}$	$\bar{\sigma}_{22}$	$\bar{\sigma}_{33}$	$\bar{\sigma}_{12}$	$\bar{\sigma}_{13}$	$\bar{\sigma}_{23}$
Tension	X	1.1715e + 02	5.8548e - 03	2.8693e - 05	6.9091e - 03	5.3393e - 03
	Y	-4.6613e - 03	1.1770e + 02	5.6218e - 03	1.1592e - 02	4.4307e - 04
	Z	1.5965e - 02	3.8249e - 03	1.1132e + 02	-1.8502e - 02	-5.7823e - 03
Shear	XY	-3.0888e - 02	-1.6751e - 02	-7.9883e - 02	2.1888e + 01	-1.8264e - 02
	XZ	-2.4071e - 02	-9.5225e - 03	-4.2720e - 02	-2.7222e - 02	2.5440e + 01
	YZ	-2.8429e - 02	-1.7819e - 02	-8.1874e - 02	-1.0655e - 02	-1.1786e - 02
						1.6665e - 03
						6.7174e - 03
						-8.3840e - 03
						-5.0730e - 03
						-8.3064e - 03
						2.5468e + 01

FEM-simulation in the component scale is performed using a multiplicative decomposition of the deformation gradient and a special form of the material law. The material law contains a submodel for the determination of the deformation behavior of the grains at each integration point using a crystal plasticity model.

The crystal plasticity model, as presented in [3], accounts for plastic deformations arising from crystallographic slip at the lattice level and also for lattice rotation due to the deformation itself. For fcc and bcc metals, the crystal plasticity model considers 12 and 48 slip systems, respectively. The number of slip systems is a direct effect of the main orientations present in fcc and bcc materials.

Once the amount of plastic deformation in the crystal scale has been homogenized, the elasto-plastic simulation in the first scale is performed. This is done with the usual predictor–corrector strategy (cf. Sect. 9.1.2), where the component scale variables for the elastic deformation gradient, plastic deformation gradient, and stress tensor are updated. The inputs for the update algorithm are the results from the crystal plasticity model, where the main variables are the shear rate and the plastic velocity gradient.

9.2 Hybrid Simulation

Jonathan Montalvo-Urquiza, Pavel Bobrov and Werner Brannath

9.2.1 Stochastic Model for Textured Metals

The mathematical and numerical investigation of material anisotropy and its effects on the elasto-plastic behavior of polycrystals has been a topic of interest for about 100 years [3]. Recent research aims at understanding the effect of stochastic fluctuations of the micro-structure in small scale specimens. For quasi-isotropic material, the averages of elastic constants can be approximated by analytical expressions [8]. However, the forming process usually induces a texture and leads to an anisotropic distribution of crystalline orientations. The characterization of anisotropic polycrystals requires numerical solutions which are usually based on Finite Element (FE) Methods [9].

This and the following sections describe how the statistical distribution of elastic properties of anisotropic micro components can be incorporated into FE methods. The focus here is on methods that account for the effects of the stochastic fluctuations in the micro-structure of the material, while external process conditions like tension and shear forces are assumed to be constant. Methods that also account for fluctuations in external input parameters like those in [10] have not yet been considered in the context of micro cold forming.

A polycrystal consists of monocrystalline grains that can vary in their orientations. If all orientations are equiprobable then the polycrystal is called quasi-isotropic. But for most specimens the orientations show a concentration around

basic directions due to the rolling of material during the production of metal sheets. In this case the polycrystal material has a texture. The texture of rolled steel and many other materials consist of fibers [11]. Each fiber corresponds to a finite set of basic orientations that arise from the production process (rolling, recrystallizations, etc.) (cf. Sect. 3.1). The orientations of the monocrystallites do not perfectly follow the basic orientations, but are statistically distributed around the directions of the texture.

The orientation of a monocrystal in the polycrystal specimen can be described by the rotation which maps the coordinate system of the specimen on the monocrystalline coordinate system. This rotation can be mathematically modeled as

$$\tilde{g} = gg_0, \quad (9.21)$$

where g_0 represents the basic directions determined by the material's production process (e.g., the orientation of the fibers resulting from a rolling process on metal sheets) and g is a random small rotation describing the stochastic variation in the grain orientation. The orientations' statistical distribution is of importance in micro-metal specimens which can consist of only few monocrystalline grains in at least one direction of the specimen. The statistical fluctuations of the orientations then have an effect on the elastic and plastic properties of the sample.

The elastic properties in each point of the polycrystalline material depend on the orientation of the monocrystal at this point [12]. We denote here the stiffness matrices in the single-grain reference system and in the global reference system by c and C , respectively. These two matrices are related as:

$$C = gg_0c, \quad (9.22)$$

with g and g_0 as defined above. Hence, the statistical distribution of the orientations g causes a statistical distribution of the stiffness tensor C .

To build a stochastic model for the specimen we need to determine the statistical distribution for the orientations. Assuming that the statistical fluctuation g of the grain orientations arises from many small and independent perturbations during the production process, a normal distribution seems to be a reasonable approximation. This normal approximation is based on Parthasarathy's central limit theorem [13]. The limiting normal distribution is not the usual normal distribution on the real numbers. It is a distribution on the rotation group. More information about this type of normal distributions can be found in [14].

According to Parthasarathy's central limit theorem, the normal distribution on the rotation group can be approximated by a large number of identically distributed and independently acting small stochastic rotations. One can, for instance, follow [14] and use small rotations that are uniformly distributed on the subsets.

$$\Pi(a, b) = \left\{ (\theta_1, \theta_2, \theta_3) : \theta_2 \leq \frac{a}{\sqrt{k}}, \quad |\theta_1 + \theta_3| \leq \frac{b}{\sqrt{k}} \right\} \quad (9.23)$$

of the space of Euler angles $(\theta_1, \theta_2, \theta_3)$. Here k is the number of small independent rotations applied. The parameters a and b describe the stochastic spread of the orientations around the basic direction. Given a and b , the distribution of crystallite orientations can be determined by a Monte-Carlo simulation. It is sufficient to simulate the distribution for a single basic direction. For other basic orientations the distribution can be obtained by rotating the generated distribution towards the actual direction.

The distribution of the orientations from the stochastic simulation can be compared to the real orientations of the crystallites in the polycrystalline by pole figures and the orientation distribution functions (ODFs) [11]. The pole figures of a polycrystal can be obtained by X-ray diffraction (XRD) [9]. As an example, here we consider DC01. A similar investigation for aluminum can be found in [7]. The analysis of an empirical ODF from experiments for DC01 shows that for this material the orientations are dominated by α - and γ -fibers [15]. Orientations from other fibers are barely visible. The pole figure from the stochastic simulation with α - and γ -fibers is shown in Fig. 9.5. The empirical distribution of DC01 was found to be best reproduced with a 7:3 ratio between the α - and γ -fibers.

For a more precise fitting of the model parameters to the empirical data, the first ODF can be used, taking the angle values $\theta_1 = 0^\circ$, $\theta_3 = 45^\circ$. This corresponds to the density plot shown in Fig. 9.6. The plot shows how both fibers are well represented by comparing the empirical and simulated distribution of the Euler angle θ_2 . The θ_2 angles at which the α - and γ -fibers occur are denoted by vertical lines.

Using the parameters derived from the comparison between simulated and observed pole figure and ODFs, one can generate the distribution of the stiffness or compliance tensor from the distribution of the orientations in Eq. (9.22).

Fig. 9.5 3D plot of the simulated 110-pole figure for steel DC01 with two fibers. *RD* and *TD* denote the rolling and traversal directions, respectively

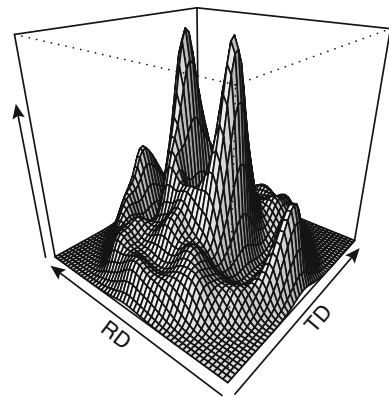
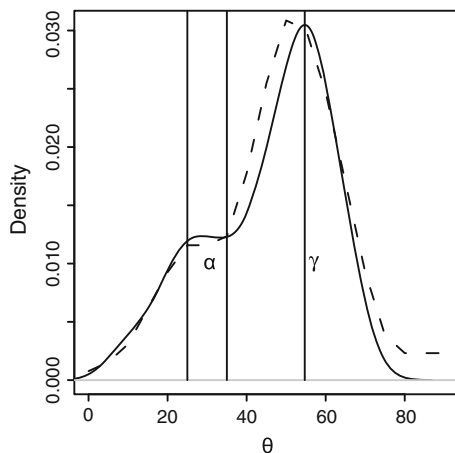


Fig. 9.6 Distributions from the stochastic simulation (solid line) and XRD-method (dotted line) of the Euler angle θ_2 when $\theta_1 = 0^\circ$ and $\theta_3 = 45^\circ$. Vertical lines indicate different components of the fibers



9.2.2 Stochastic Two-Scale FEM

The stochastic model from Sect. 9.2.1 can be used in a numerical simulation for a component which is produced by micro metal forming. The goal is to account for the effects of the stochastic fluctuations in the orientation and geometry of single grains in the component.

Using the ideas of the two-scale methods presented in Sect. 9.1, it is possible to implement a two-scale method for the simulation of material response including some stochastic models in the second scale [12]. In this context, the first scale corresponds to the whole component under consideration and the second one corresponds to the scale of the RVE. In computational mechanics literature the first and second scales are referred to as “macro” and “micro” scales, respectively. These names are not used here to avoid confusion in the context of micro components.

The main difference to a conventional two-scale FEM is the stochastic nature of the material properties inside the representative volume element (RVE) [12]. The aggregate of several grains forming the RVE, as defined in Sect. 9.1, includes two main sources of randomness which must be considered: the geometric shapes of the mono-crystalline grains and the orientations of the single-grain lattices. Anisotropies arising from these random sources are computationally tracked with the numerical meshing for the RVE geometry, and the varying stiffness tensors for the different grains obtained from the stochastic texture model.

The stiffness matrix C in the RVE domain with n grains can be described as a function of the position inside the domain (cf. Sect. 9.1). Using the texture model from the previous section, this function is:

$$C(x) = C(j), \quad j = 1, \dots, n, \quad (9.24)$$

where j represents the index of the grain and x is a point inside the domain, i.e., $x = (X, Y, Z)$. Each matrix $C(j)$ can be described with the orientation of the corresponding fiber component and a random small rotation, as explained in [Sect. 9.2.1](#). With this stiffness, the values of stress at any point in the RVE can be computed using Hooke's law as $\sigma(x) = C(x)\varepsilon(x)$ and then homogenized with the integrals of the form

$$\bar{\sigma}_{ij} = \frac{1}{|RVE|} \int_{RVE} \sigma_{ij}(x) dx, \quad i, j \in \{1, 2, 3\}. \quad (9.25)$$

Computing and storing the homogenized tensors can describe the individual response of a RVE when loads are applied. The randomness of grain geometries and lattice orientations are captured by the stochastic texture model and the random generation of crystal centers.

Via repeated computations on different randomly generated material realizations, we arrive at what we call a Material Library. This provides a discrete approximation of all possible responses of a material under elastic loads. The idea of this library is that every computation in the homogenization step of a two-scale FEM method can be easily obtained by mapping a given strain tensor into a volume averaged stress which now has a stochastic form, instead of a deterministic one as the classical FE^2 method presented in [Sect. 9.1](#).

With a Material Library, every stress calculation in the RVE scale is reduced to a simple additive evaluation in the library, avoiding the expensive calculation of a finite element simulation for every integration point in the component scale. To use the Material Library, it is only necessary to split the given strain into its components as:

$$\varepsilon = \varepsilon_{xx}e_1 + \varepsilon_{yy}e_2 + \varepsilon_{zz}e_3 + \varepsilon_{xy}e_4 + \varepsilon_{xz}e_5 + \varepsilon_{yz}e_6 \quad (9.26)$$

and to calculate the stochastic values of stress by multiplying each strain value with the corresponding distribution in the Material Library.

The diagram in [Fig. 9.7](#) summarizes the steps for constructing a Material Library for texturized materials. This Material Library starts from the stochastic nature of the single-grain stiffness determined by the main texture components, includes the randomness in the RVE geometry, and ends with a complete stochastic description of the mechanical properties for the component scale.

The computational effort to create the material library with N different material realizations is equivalent to the effort to solve one FE^2 method with $6N$ quadrature points. Furthermore, the method has the important advantage that the material library only needs be computed once for a given material, regardless of how many simulations must be performed.

Material Libraries created with about one thousand realizations have been shown to include enough stochastic information about the material properties. This makes the construction of the library very affordable. The computational effort is similar to the one needed for a single FE^2 simulation in a domain with one

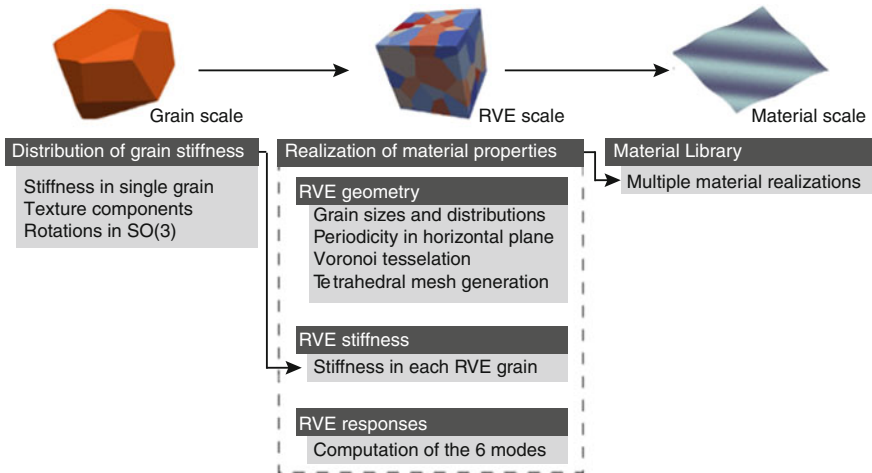


Fig. 9.7 Diagram for the construction of a Material Library for texturized material and periodic RVEs

thousand degrees of freedom, which is a very small number when considering three-dimensional problems.

For the case of a DC01 foil with a thickness of $50\ \mu\text{m}$, the main components of the Material Library obtained from 1,200 material realizations are presented in Fig. 9.8 in the form of histograms. From these histograms, some of the usual material responses for a foil can be observed. For example, the tensile response in the foil's horizontal directions (X and Y) is very similar, while the response in the direction of thickness (Z direction) is of a different size. It can also be observed how the absolute size of the stress differs between shear and tension modes, the tension mode being the one creating higher levels of stress. However, the amount of tensile stress is in many cases not decisive for permanent (plastic) deformations, while the shear stresses are directly related to such effects.

9.2.3 Application of the Stochastic Two-Scale Method to the Bulging Test

The use of the Material Library is illustrated here by showing the different stress responses at some points during a bulge test on a circular piece of DC01 steel foil with a thickness of $50\ \mu\text{m}$ and radius of $450\ \mu\text{m}$.

Figure 9.9 shows the shape obtained after simulating the process. This simulation has been performed by computing a FEM solution for the displacement and strain fields using the expectation of the stiffness matrix (also taken from the Material Library). In Fig. 9.9b, the colors represent the computed expectation of

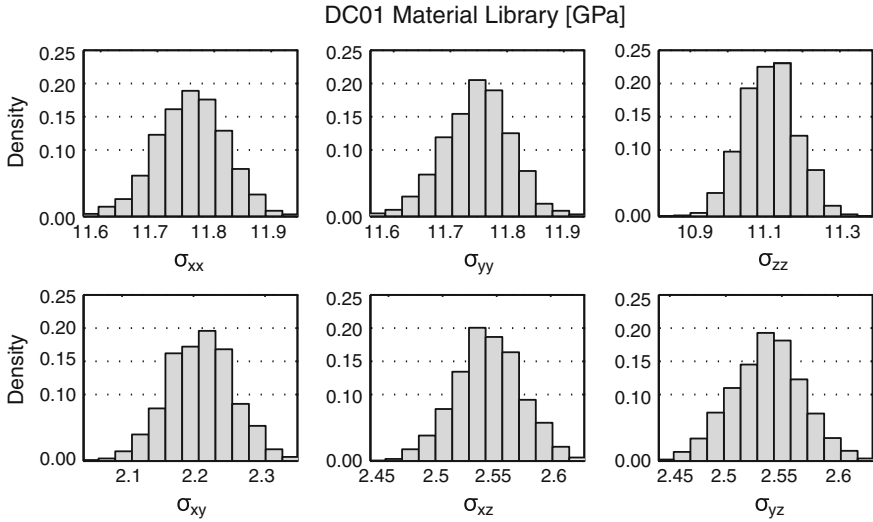


Fig. 9.8 Material Library presented in the form of material response distributions. Each *histogram* shows the distribution of stress (GPa) for loads of 1 % in each of the tensile (*upper row*) and shear modes (*lower row*)

the von Mises stress. The anisotropies inherent to the texturized material can be observed in radial equivalent areas with different stress results.

In Fig. 9.9, three specific points have also been chosen on the foil piece, one at the top, one close to the edge, and one in the middle. The selected points belong to three different areas of low stress (point 1), medium stress (point 2) and high stress (point 3). The stochastic distributions of the stress tensor at these selected points have been calculated using the strains at the component scale and the pre-computed Material Library. The stress distributions can be found in Fig. 9.10.

It is easy to note that the selected locations correspond to very different levels of stress. This can also be confirmed by analyzing the values in the horizontal axes of the subplots in Fig. 9.10. One can observe, for example, the very different sizes on the tensile stresses for Point 3 (cf. Fig. 9.10c) which together with the large shear in the XY and XZ planes cause the largest equivalent stress among the three points. The Points 1 and 2 (cf. Fig. 9.10a, b) show also unequal values for the different stress components, but are much smaller than the ones for Point 3.

Note that here the importance of the stochastic modeling and simulation is not only to provide information on the stress levels that can be expected in the component, but also to determine local spreading of such values through the distributions obtained. In this form, predictions about critical values of stress can be obtained in the form of probabilities for any desired location in a simulated component. This can be very convenient at design stages and in the prototyping of new micro components. A simulation performed in this way can be used to provide a map of probable high stresses in a component, either for the whole piece or for a

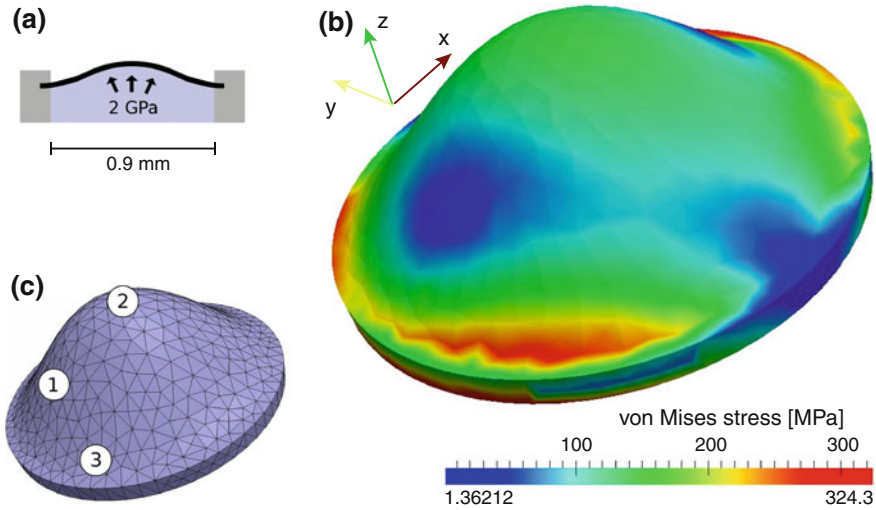


Fig. 9.9 **a** Bulge test setting simulated for DC01 foil with thickness of 50 μm. **b** Final shape (deformation scaled by a factor of 20) and expectation of the von Mises stress. **c** Points with low (1), medium (2), and high (3) equivalent stress for the analysis of stochastic local stresses, shown in Fig. 9.10

region within it. With this, variations in the micro component's structure can be tracked without the need of simulating each possible production sample.

Regarding the computational effort, the method presented here allows the computation of both deterministic and stochastic values without huge computational costs. Calculations on component scale are kept at the same complexity level as in any standard simulation, while calculations in the second scale are negligible once the Material Library is produced [12].

The local distributions of stress provide evidence of the existence of areas in which the points might reach plastic deformation in terms of probability. This idea can be used to later describe not only plastic strains and stresses, but also the probability that they appear. The construction and use of Material Libraries for the plastic case is a current research topic.

A major issue in this direction is the handling of stochastic distributions in the one-time step of elasto-plastic computation (cf. Sect. 9.1). The equilibrium equation to be solved in the predictor step is deterministic but must now account for the plastic strain distribution. In principle this could be done by Monte-Carlo simulations. However, a Monte-Carlo approach becomes infeasible when applied to a large number of time steps with a large number of degrees of freedom. Taking a simple summary measure for the plastic strain distribution (as e.g., the expectation) might not sufficiently capture the stochastic nature of the problem.

Another possible approach is to connect the stochastic two-scale FEM with the Crystal Plasticity method (cf. Sect. 9.1) by finding a set of basis functions for the

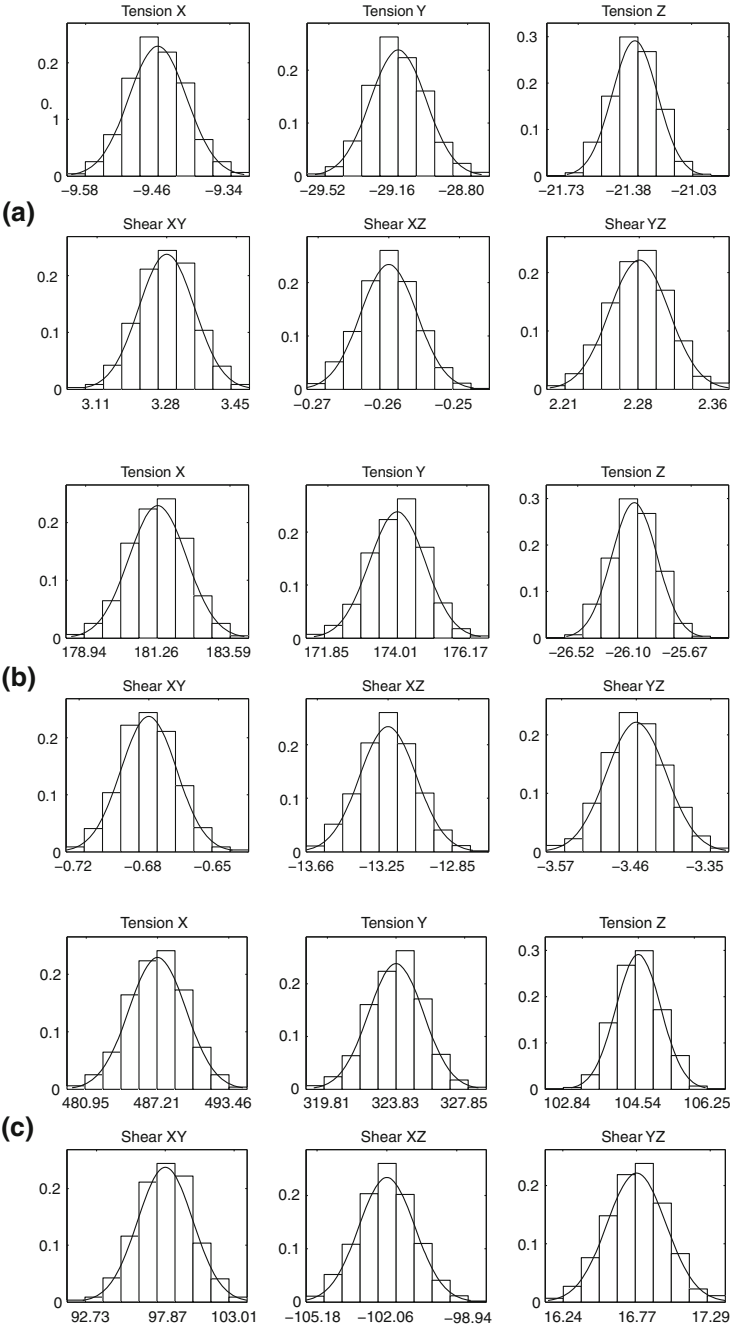


Fig. 9.10 Distributions for local stress tensors at selected locations from Fig. 9.9a–c

slip systems at the crystal level. With this idea, a modified construction of the RVE responses could be obtained and combined for the construction of a plastic library.

9.3 Inverse Modelling

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The development and manufacturing of high quality micro machined structure parts goes together with modeling, simulation, and optimization. In order to obtain the optimal manufacturing results or to determine the causes of any unsatisfactory manufacturing results, efficient models and simulations depending on the structures and the process parameters are needed. To analyze the mathematical methods for the parametrization and optimization of simulation models, two things are needed. There has to be developed a forward model, which will depend on the particular system, structures, material parameters, and process parameters, e.g., feed rates, depth of cut, stiffness, and damping parameters for the machine structures. Secondly, an appropriate inverse model is required. In real processes, only a rough approximation of some of these parameters might be available or they might not be known at all. The relation between the parameters q and the measurements m is mathematically described by the operator equation:

$$Q(q)=m, \quad (9.27)$$

which is usually called “the parameter to state mapping”. This model allows simulating the measurements m for different parameters q . The process parameters q can be constants or functions depending on time and/or space. The approaches used to fit the model to the data are the so-called inverse modeling methods. These techniques allow finding the feasible values of the parameters q . In the case of constant parameters, one treats of the calibration of the model, and Eq. (9.27) is usually well-posed. If the parameters $q = (t, x)$ are functions, then parameter identification methods need to be employed.

$$\min_q \|Q(q) - m\|_2. \quad (9.28)$$

However, only a noisy version m^δ of the true data and the noise level δ with

$$\|m - m^\delta\|_2 \leq \delta \quad (9.29)$$

are known. In general, the problem (9.28) is ill-posed, meaning either that the operator Q is not surjective on the space of all noisy data, that the problem does not have a unique solution, or that the inverse Q^{-1} is not continuous. In such a case, the so-called regularization method has to be applied to determine the parameters q . Probably the best known regularization technique for linear as well

as nonlinear inverse problems is Tikhonov regularization [16]. Here, the solution is obtained by minimizing the Tikhonov functional,

$$\min_q \|Q(q) - m^\delta\|_2^2 + \alpha \Phi(q), \quad (9.30)$$

where Φ is a non-negative penalty function suitably chosen to incorporate a priori information about the true solution. Often the penalty functional,

$$\Phi(q) = \|q\|_w = \left(\int |q(\xi)|^w d\xi \right)^{\frac{1}{w}}, \quad (9.31)$$

for $0 < w < \infty$ is used. Here, $\|\cdot\|_w$ for $0 < w \leq \infty$ means the norm of the Lebesgue spaces $L_w(\mathbb{R})$, usually denoted by $L_p(\mathbb{R})$. When $w = 2$, the techniques of Hilbert spaces are used, which have been widely investigated in recent years, see [16]. The presented parameter identification techniques are illustrated on an example for the manufacturing process of micro parts. The quality of the product obtained by micro forming depends on the tribological properties of the micro molds. Hence, mathematical models are needed for describing and characterizing its tribological surface properties. The micro milling operations are taken into account. The model development is based on two steps: numerical simulation for computing the actual tool tip position, and the inverse method for determining the optimal input parameters for the required surface quality.

9.3.1 Forward Model

This section will discuss the model development of the tool tip position for micro milling operations. Only the main parts of the model will be given (for further information, see [17]). The aim is the development of a model to characterize the structures of the cutting processes with a defined cutting edge. The model includes the dynamical changes of the tool caused by the forces occurring during the process. These changes have a direct impact on the removal of the material, and hence on the machined surface. The interaction between the tool and the workpiece is described by a system of ordinary differential equations (ODEs). The solution of this system provides the actual parameters and forces. Changes of parameters lead to unexpected differences in the positions of the tool and thereby in the surface generated. On the other hand, in the micro-range, size effects have to be included, see [18].

The ball end milling cutter which is rotating and moving along the given path through the workpiece [19] is analyzed. The tool moves along the x -axis with a feed velocity v_f and cuts the workpiece with a depth of cut a_{pc} . The first part of the model describes the position of the tool tip at each process time t . One of the major factors causing dynamical changes in the relative work–tool position are the forces acting during the process. It is therefore essential to predict these forces precisely.

In developing a dynamical model and calculating the actual model parameters, the process parameters are defined to be time dependent functions, which can be integrated later into the tool position model.

A coordinate system where the feed motion v_f of the tool is in the x direction and the z coordinate indicates the depth of cut a_{pc} is introduced, cf. Fig. 9.11. The tool tip position at time t is described by a vector $(x(t), y(t), z(t))^T$ and is influenced by different effects. We take into account the run-out η and the deflections $\delta_r(t), \delta_t(t), \delta_{ax}(t)$ of the tool in the radial, tangential, and axial directions, see Fig. 9.12. For each direction $i = \{r, t, ax\}$, they are computed as the fraction of the force $F_i(t)$ over the stiffness c_i by:

$$\delta_i(t) = \frac{F_i(t)}{c_i}. \quad (9.32)$$

By applying the coordinate transformation

$$\Omega(\mathbf{t}) = \begin{pmatrix} \cos(\varphi(t)) & -\sin(\varphi(t)) & 0 \\ \sin(\varphi(t)) & \cos(\varphi(t)) & 0 \\ 0 & 0 & 1 \end{pmatrix}, \quad (9.33)$$

the deflections $\delta_x(t), \delta_y(t), \delta_z(t)$ on the (X, Y, Z) system are obtained. Moreover, as an additional effect, the total tool deflection is considered. It is calculated by means of the resulting force $\mathbf{F}(\mathbf{t}) = \mathbf{F}_t(\mathbf{t}) + \mathbf{F}_r(\mathbf{t})$ such that the variation of the depth of cut is given by

$$a_{pc}^{var} = \frac{\delta^2(t)}{2l_h} = \frac{\delta_t^2(t) + \delta_r^2(t)}{2l_h}, \quad (9.34)$$

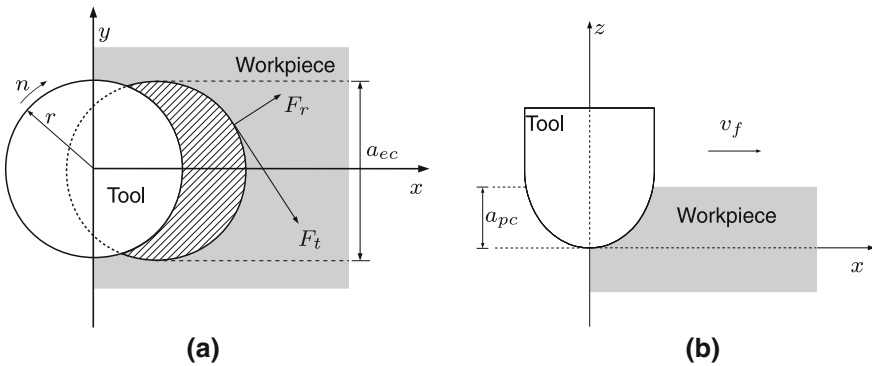


Fig. 9.11 Coordinate system in milling operations. **a** (X, Y) -plane **b** (X, Z) -plane

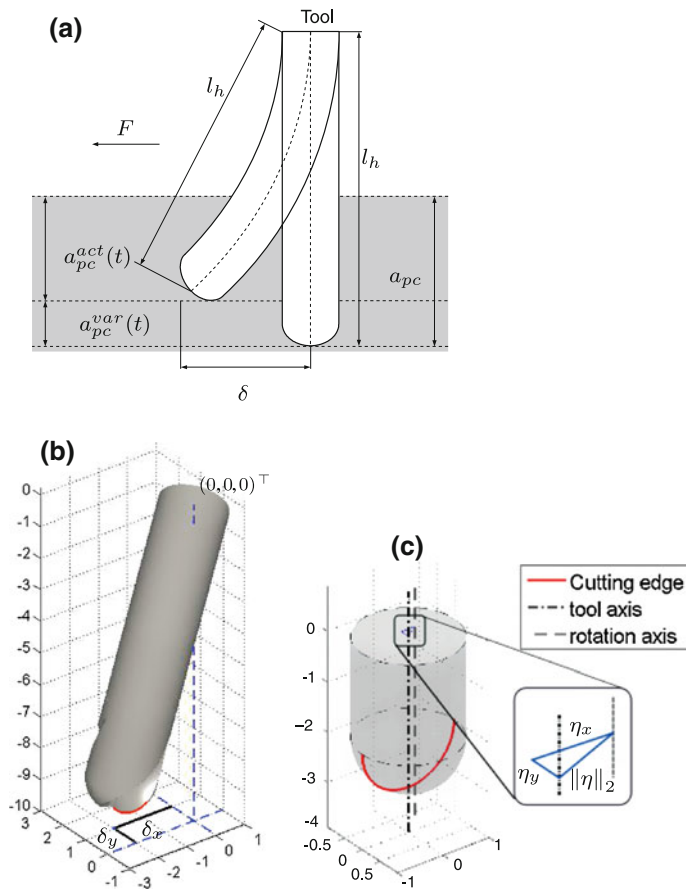


Fig.9.12 Dynamic deviations in the relative work-tool position. **a** Variation of the depth of cut. **b** 3-dimensional view of the tool deflection with $\delta_x = -2, \delta_y = 1$. **c** Model of a 2-fluted ball-end tool, cutting edge, and the rotation axis with parallel offset

where l_h denotes the length of the deflected tool. For more details, see the approach developed in [20]. A similar method has been developed for precision turning process, see [21].

The rotational angle is given by $\varphi(t) = 2\pi nt$, where n is the rotational speed of the tool. In a first modeling step, the acting forces, which can be separated into three components, namely the radial force $F_r(t)$, the tangential force $F_t(t)$, and the axial force $F_{ax}(t)$, are considered. The vector of these force components is denoted by

$$\mathbf{F}^*(t) = (F_r(t), F_t(t), F_{ax}(t))^T. \quad (9.35)$$

The force model is based on the usual approach in which the force is proportional to the chip thickness $h_c(t)$ and the axial depth of cut $a_{pc}(t)$. It can be extended by a second term which describes the frictional force, see [22]. This tool position model is flexible and can be easily generalized, for instance by taking into account other models for the micro forces. Furthermore, let j be the number of cutting edges. Thus, the total force at an arbitrary cutter rotation angle is evaluated by adding the forces acting on each cutting edge, i.e.,

$$\mathbf{F}^*(t) = \sum_{i=1}^j \mathbf{F}^{*,i}(t) = \sum_{i=1}^j \mathbf{B}_k^i a_{pc}(t) h_c^i(t) + \mathbf{B}_m^i a_{pc}(t), \quad (9.36)$$

where $\mathbf{B}_k^i = (B_{k,r}^i, B_{k,t}^i, B_{k,ax}^i)^T$ and $\mathbf{B}_m^i = (B_{m,r}^i, B_{m,t}^i, B_{m,ax}^i)^T$ are the vectors of cutting constants which are dependent on the material properties and have to be derived from measurements. By applying the coordinate transformation (9.33), the resulting forces in the three orthogonal directions $\mathbf{F}(t) = (F_x(t), F_y(t), F_z(t))^T$ of the (X, Y, Z) system are obtained. Furthermore, the chip thickness is

$$h_c(t) = \frac{v_f^{act}(t)}{nj} \sin(\varphi(t)). \quad (9.37)$$

In the second step, the forward model is extended by a model for the actual process parameters and the actual tool path. Here, the approach introduced in [20] is followed. The tool path at the time t in the coordinate system (X, Y, Z) is then given by

$$\begin{aligned} x(t) &= \int_0^t v_f(\xi) d\xi - \delta_x(t) + \eta_x(t), \\ y(t) &= -\delta_y(t) + \eta_y(t), \\ z(t) &= a_{pc}(t) - \frac{\delta_r^2(t) + \delta_t^2(t)}{2l_h} - \delta_z(t). \end{aligned} \quad (9.38)$$

Note that δ_z equals δ_{ax} because of the transformation (9.33). Moreover, η_x and η_y are the components of the static run out vector $\eta = (\eta_x, \eta_y, 0)$ (see Fig. 9.12), and for the run out angle γ they are computed by

$$\eta(t) = (\|\eta\|_2 \sin(\varphi(t) + \gamma), \|\eta\|_2 \cos(\varphi(t) + \gamma), 0)^T. \quad (9.39)$$

For a given input depth of cut $a_{pc}(t)$ and input feed velocity $v_f(t)$, $z(t)$ can be interpreted as the real depth of cut $a_{pc}^{act}(t) := z(t)$ and $\dot{x}(t) := v_f^{act}(t)$ as the real feed velocity $v_f^{act}(t)$. Combining Eq. (9.38) describing the tool path at the time t and Eq. (9.34) modeling the deflections of the micro tool with the force model (9.36) and their derivatives, one obtains the system of ordinary differential equations

$$\begin{aligned}
\dot{\varphi}(t) &= 2\pi nt, \\
\dot{x}(t) &= v_f(t) - \dot{\delta}_x(t) + \dot{\eta}_x(t), \\
\dot{y}(t) &= -\dot{\delta}_y(t) + \dot{\eta}_y(t), \\
\dot{z}(t) &= \dot{a}_{pc}(t) - \dot{\delta}_{ax}(t) - \frac{\dot{\delta}_r(t)\dot{\delta}_t(t) + \dot{\delta}_r(t)\dot{\delta}_r(t)}{l_h}, \\
\dot{h}_c(t) &= \frac{v_f^{act}(t)}{nj} \cos(\varphi(t)) \dot{\varphi}(t) + \frac{\dot{v}_f^{act}(t)}{nj} \sin(\varphi(t)), \\
\dot{F}_i(t) &= B_{k,i} \dot{a}_{pc}^{act}(t) h_c(t) + B_{k,i} \dot{a}_{pc}^{act}(t) \dot{h}_c(t) + B_{m,i} \dot{a}_{pc}^{act}(t), \quad (i = r, t, ax) \\
\dot{\delta}_i(t) &= c_i^{-1} \dot{F}_i(t), \quad (i = r, t, ax) \\
\dot{\delta}_x(t) &= -\sin(\varphi(t)) \dot{\varphi}(t) \delta_r(t) + \cos(\varphi(t)) \dot{\delta}_r(t) - \cos(\varphi(t)) \dot{\varphi}(t) \delta_t(t) - \sin(\varphi(t)) \dot{\delta}_t(t), \\
\dot{\delta}_y(t) &= \cos(\varphi(t)) \dot{\varphi}(t) \delta_r(t) + \sin(\varphi(t)) \dot{\delta}_r(t) - \sin(\varphi(t)) \dot{\varphi}(t) \delta_t(t) + \cos(\varphi(t)) \dot{\delta}_t(t)
\end{aligned} \tag{9.40}$$

for the special case of $j = 2$. Its solution yields the actual process parameters as well as the actual forces at each time step. With these results, we are able to compute the resulting position of the tool tip. This gives a forward model

$$(x(t), y(t), z(t)) = Q(a_{pc}(t), v_f(t)), \tag{9.41}$$

where Q is the operator solving the system of ODEs for the input parameters $a_{pc}(t)$ and $v_f(t)$.

9.3.1.1 Numerical example of the forward model

MATLAB routines were used to solve the system of ODEs numerically. The solution is the vector of the actual process parameters and the actual forces at each time of the process. The model was validated in order to derive the parameters which are consistent with the measured cutting force. For the comparison of the outcomes of the model simulations with the measurements, we refer to [17]. The example was calculated using, as input parameters, the depth of cut $a_{pc}^0(t)$ and feed velocity $v_f^0(t)$ as plotted in Fig. 9.13. The input feed rate as well as the input depth of cut is reduced by the elastic deflection of the tool due to the cutting forces appearing during the process. In Fig. 9.14, the positions $z(t)$ and $x(t)$ computed by the forward model for these input parameters are showed as dashed lines. They are compared with the input depth of cut and then the ideal position $x^0(t)$ which corresponds to the input feed velocity. The difference in both curves is shown.

9.3.2 Parameter Identification Problem in Micro Cutting Operations

An inverse parameter identification problem for cutting processes is introduced and solved by regularization methods with sparsity constraints. The forward model

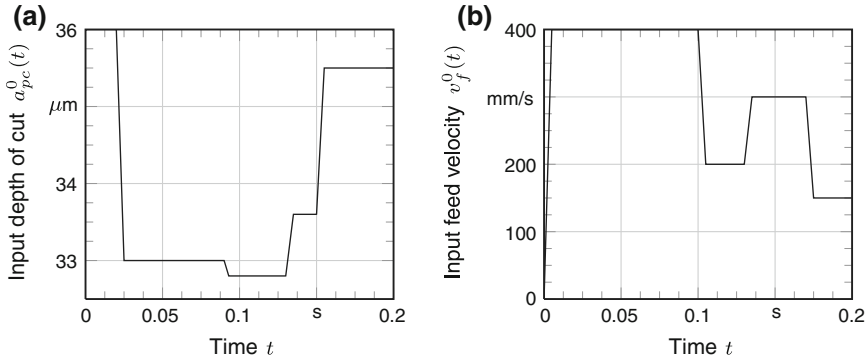
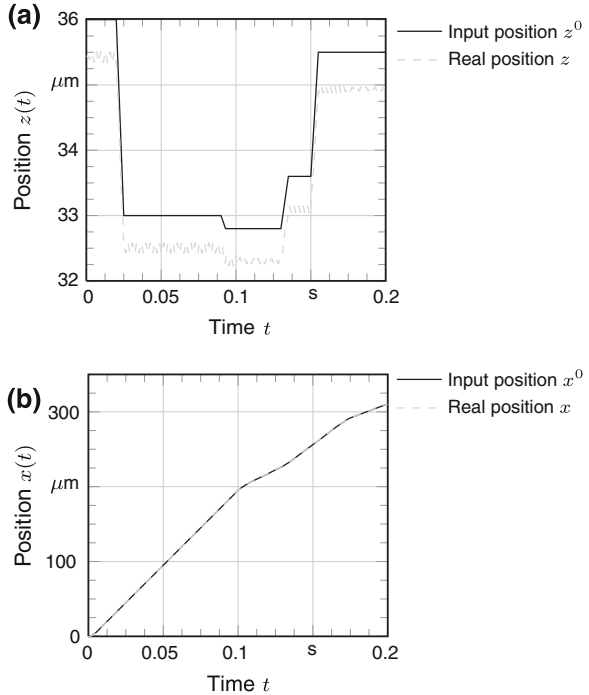


Fig. 9.13 Given input parameters. **a** Input depth of cut $a_{pc}^0(t)$. **b** Input feed velocity $v_f^0(t)$

Fig. 9.14 Resulting

positions $(x(t), z(x)) =$

$Q(a_{pc}^0(t), v_f^0(t))$ computed by the forward model for input parameters a_{pc}^0 and v_f^0 from Fig. 9.13. **a** Given depth of cut $a_{pc}^0 = z^0$ and computed vertical position z . **b** Given position $x^0 = \int v_f^0(\xi) d\xi$ and computed position



(9.41) which allows computing the tool tip position $m = (x(t), z(t))$ for different choices of parameters $q = (a_{pc}, v_f)$ is used. Moreover, only noisy data $m^\delta = (x^\delta(t), z^\delta(t))$ is given, see Fig. 9.15. To obtain the best setting of the parameters $q = (a_{pc}, v_f)$, the minimizing problem (9.30) has to be solved. The problem under consideration is ill-posed, so that appropriate regularization methods have to be

applied. Because of restrictions on the cutting machine, we search for a solution to the problem (9.28) which has a sparse representation, i.e., the solution can be expressed in terms of an orthogonal basis or frame $\zeta := \{\zeta_i\}_{i \in \mathbb{Z}}$ by

$$q = \sum_{i \in \mathbb{Z}} \langle q, \zeta_i \rangle \zeta_i, \quad (9.42)$$

where only a finite number of coefficients $q_i = \langle q, \zeta_i \rangle$ are nonzero. To achieve this, the sparsity functional (9.31) is taken with $w = 1$ in the penalty term Φ , i.e.,

$$\Phi(q) = \int |q(\tau)| d\tau. \quad (9.43)$$

These optimization techniques have recently become available. Especially, the method presented in [23] is a breakthrough in the theory of inverse problems and their applications.

Based on the method developed in [24], analogous techniques for micro milling processes will be presented. The notations given there are used here, and, since the methods are very similar, only the general idea without any focus on the details will be presented.

Assume that the positions $x(t)$ and $z(t)$ are given. From (9.40) it follows that $\delta_i(t)$ for $(i = r, t, ax)$ and $\delta_x(t)$ can be written as a function of $v_f^{act}(t) = \dot{x}(t), x(t)$ and $z(t)$. Together with (9.38), we have

$$\int_0^t v_f^0(\xi) d\xi = x(t) + \delta_x(x, \dot{x}, z) \quad (9.44)$$

and

$$a_{pc}^0(t) = z(t) + \frac{\delta_r^2(x, \dot{x}, z) + \delta_t^2(x, \dot{x}, z)}{2l_h} + \delta_z(x, \dot{x}, z) \quad (9.45)$$

We refer to these functions as $m_1(t)$ and $m_2(t)$, respectively. For the integral operator

$$\mathcal{A}_1 : v \rightarrow \int v(\xi) d\xi \quad (9.46)$$

and the identity operator

$$\mathcal{A}_2 : v \rightarrow v, \quad (9.47)$$

Equations (9.44) and (9.45) can be reformulated in terms of the general setting of (9.27):

$$\mathcal{A}_1(v_f^0) = m_1 \quad (9.48)$$

and

$$\mathcal{A}_2(a_{pc}^0) = m_2. \quad (9.49)$$

The regularization methods with sparsity constraint, discussed above, for computing the best input parameters will now be applied. This example covers the problem of computing the optimal input parameters in order to achieve the desired height of the ramp. The goal is the computation of the optimal depth of cut a_{pc} for a given height of the ramp. Therefore, the Haar wavelet basis will be used for the representation of the solution.

Assume that a_{pc}^1 and a_{pc}^2 are given and that $a_{pc}^1 > a_{pc}^2$. The milling tool starts to cut with a given depth of cut a_{pc}^1 and at a given time, t_1 , the depth of cut changes to a_{pc}^2 . In Fig. 9.15, we see the measurements of such an experiment and its resulting depth of cut. The desired height difference $\Delta a_{pc} = a_{pc}^1 - a_{pc}^2$ was $20 \mu\text{m}$. The measured values are $a_{pc}^1 = 36.241 \mu\text{m}$ and $a_{pc}^2 = 15.393 \mu\text{m}$, so that the real difference is $\Delta a_{pc} = 20.848 \mu\text{m}$. The parameters of the experiment are collected in Table 9.2.

Fig.9.15 Measured surface for a milling experiment with variation of depth of cut. **a** 3D representation. **b** Profile of the measured depth of cut

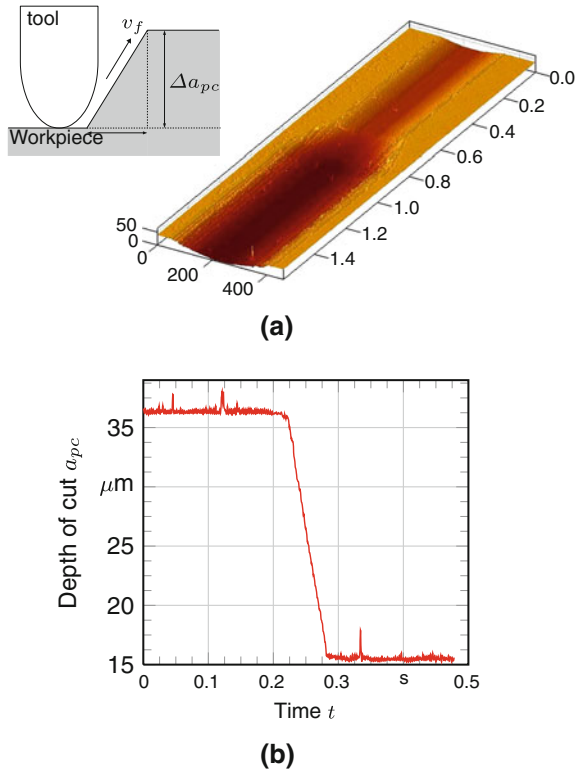


Table 9.2 Experimental milling conditions

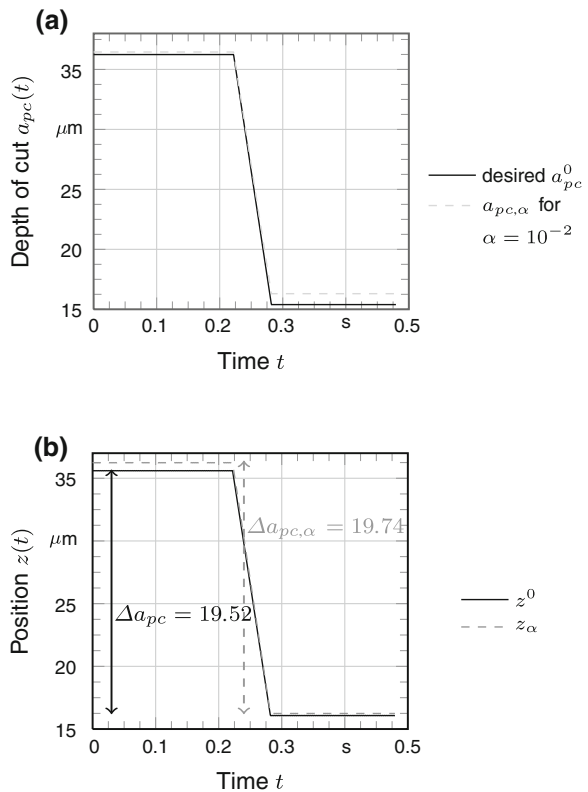
Radius of the tool	$r = 0.5 \text{ mm}$
Length of the ramp	$\Delta x = 0.2 \text{ mmZ}$
Rotation speed	$n = 20 \text{ rpm}$
Feed velocity	$v_f = 200 \text{ [mm/min]Z}$

The task is the computation of the optimal inputs $a_{pc,\alpha}^1$ and $a_{pc,\alpha}^2$ so that the resulting surface has the ramp $\Delta a_{pc,\alpha} = 20 \mu\text{m}$. The desired a_{pc}^* is plotted as a solid line in Fig. 9.16a and is used to calculate the function m_2 . Taking into account the regularization, we have to minimize the problem

$$a_{pc,\alpha} = \arg \min_{a_{pc}} \|\mathcal{A}_2(a_{pc}) - m_2\|_2 + \alpha \|a_{pc}\|_1. \quad (9.50)$$

The solution $a_{pc,\alpha}$ is plotted as the dashed line in Fig. 9.16a. We plug the a_{pc} and the optimal $a_{pc,\alpha}$ into the forward model and obtain the positions z^* and z_α , respectively. In Fig. 9.16b, the position z^* is plotted against time as the solid line

Fig. 9.16 Input and corrected depth of cut and their corresponding initial position and the corrected position. **a** Input a_{pc}^0 and corrected optimal $a_{p,\alpha}$. **b** The corresponding initial position z^0 and z_α



and compared with the position z_α plotted as the dashed line. We observe the improvement of the resulting height of the ramp.

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Chapter 10

Quality Aspects

Ralf B. Bergmann

Latin

a	Orthogonal distance of a line in R2 to the origin (mm)
b_{actual}	Actual deviation (mm)
$b_{\text{specified}}$	Tolerated deviation (mm)
C_p	Capability index
C_{pk}	Minimum capability index
C_{pkLL}	Lower capability index
C_{pkUL}	Upper capability index
d	Lateral resolution (mm)
d_{ai}	(Orthogonal) Distance of point with index i (mm)
k	Factor k depends upon the type of probability distribution of the output quantity in a measurement model and on the selected coverage probability
LL	Lower limit
n	Refractive index
NA	Numerical aperture
Pr	Corresponding probability
P_c	Cost efficiency evaluation parameter
P_p	Process performance index
P_{pk}	Minimum process performance index
P_{pkLL}	Lower process performance index
P_{pkUL}	Upper process performance index
P_q	Quality evaluation parameter
P_t	Technological evaluation parameter

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Q_G	L_2 -norm or gaussian norm (mm)
Q_{sup}	Supremum/Chebychev norm (mm)
r	Radius (mm)
$r_{\text{circumscribed}}$	Radius of a minimum circumscribed element (mm)
$r_{\text{inscribed}}$	Radius of a maximum inscribed element (mm)
r_{L2}	Radius of a Gaussian element (mm)
r_{sup}	Radius of a chebychev/minimum zone element (mm)
$U(x, y)$	Complex field amplitude (V/m)
UL	Upper limit
U	Expanded measurement uncertainty
u_i	Individual standard measurement uncertainty
u	Coordinate (m)
v	Coordinate (m)
W_c	Cost efficiency weighting factor
W_{cc}	Costs
W_{ccm}	Manufacturing costs
W_{ccs}	Setup costs
W_{cct}	Transport costs
W_{ct}	Time
W_{ctm}	Manufacturing time
W_{cts}	Setup time
W_{ctt}	Transport time
W_q	Quality weighting factor
W_{qi}	Impacts of quality
W_{qip}	Impact of process changing
W_{qir}	Impact of rechecking
W_{qm}	Quality of manufacturing processes
W_t	Technological weighting factor
W_{tr}	Resource availability
W_{ts}	Technological sensibility
X_i	Influencing input quantities
X_{LL}	Lower quantil
X_{UL}	Upper quantil
X	Coordinate (m)
x_{in}	n measured data for quantity X_i
$\bar{x}_i$	Mean value of n measured data for quantity X_i
Y	Model function
y	Coordinate (m)

Greek

α	Half angle of the aperture (°)
β	Angle (around z-axis) (rad)

γ	Tip angle (°)
Δ	Estimator for dispersion
ε_i	Corresponding expectation value
λ	Wavelength (nm)
η	Coordinate (m)
Λ	Synthetic wavelength (μm)
μ	Estimator for location of distribution
μ_{add}	Estimator for additional fluctuations of the location of distribution
ξ	Coordinate (m)
$\sigma_{\text{emp},i}$	Empirical standard deviation of the influencing quantity X_i
$\sigma_{\text{emp},y}$	Combined standard measurement uncertainty for the expectation value y of Y
σ_i	Standard deviation of the influencing quantity X_i

10.1 Metrology

10.1.1 Quality Inspection

Gert Goch and Karsten Lübke

Tactile, optical and the growing field of tomographical measurement are widely dominating the inspection of manufactured workpieces, applying a broad variety of measuring devices. These instruments measure the deviations of the specified geometry defined in the design drawings. In general, these features can be subdivided into size, form, position according to ISO 1101 and the microtopography of the measurand. Geometry measuring instruments are calibrated and thus the measured lengths are based on meters. The meter is one of the seven base units of the International System of Units (SI) [7]: Unit of length in meter. The definition is based on the speed of light travelling in a vacuum. In general, measurement is an experimental process, comparing the objective geometric measurand with the appropriate unit, i.e. the meter or a reference angle. The measurement result is compared with the tolerated specified dimensions within a quality inspection. Regarding micro metrology, manufactured workpieces offer features in different sizes. Following the definition of Geiger [20], micro features are specified by at least two dimensions in the sub-millimeter range. Therefore, manufactured parts in geometric metrology can be divided into two groups: Macro parts with macro and micro features and micro parts with micro features. Regarding micro metal forming, Fig. 10.1 shows a micro deep drawing ring and punch and a micro-cup. The micro deep drawing ring is a macro part with an outer diameter of more than 30 mm. But, the forming zone has a diameter and forming radius below 1 mm. The corresponding punch is also a macro part with a forming zone with micro features. Both tools are required to manufacture a micro-cup, which is a cold formed micro part with micro

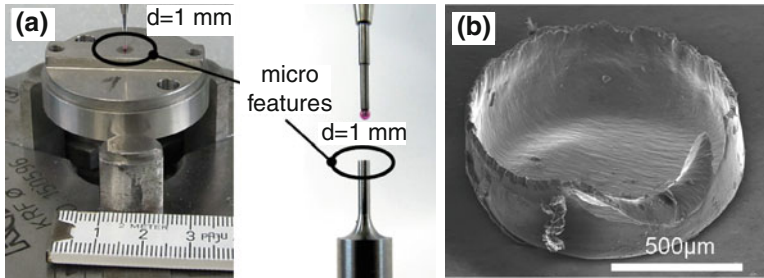


Fig. 10.1 Examples of macro and micro parts in dimensional metrology according to [20]: **a** macro parts of micro deep drawing ring and punch with macro and micro features and **b** micro cold formed micro-cup

features, e.g. a cylindrical surface. Its diameter is below 1 mm. Micro features can also be topographical structures or a specified waviness of the surface.

The quality inspection of macro parts is in general less challenging than micro parts. There are differences in the handling and clamping of macro and micro manufactured parts. Handling of manufactured parts covers temperature change, storage, transportation and changes in the surface, e.g. by oxidation.

Deviations from the specified dimensions are related to different reasons. The VDI/VDE guideline 2601 states that “Every workpiece, no matter how precisely it has been manufactured, displays deviations from the geometrically ideal form” and defines the following deviation types [38]:

- machine-dependent deviations:
 - static and dynamic deviations due to guide rails and bearings of moving machine components
 - positioning deviations of moving machine components
 - elastic deformations of machine, guide rails, tool, etc.
 - tool wear
 - bearing play
 - vibrations between the tool and the machine
- workpiece-dependent deviations:
 - material inhomogeneities
 - deformation of the workpiece during processing
 - differing local temperature distribution during the production process
 - subsequent shrinkage after processing
 - release of inner stress after processing
 - deformation due to hardening
- environmentally-dependent deviations:
 - local temperature fluctuations
 - temporal temperature fluctuations
 - vibrations transferred to the machine from the surroundings via the foundation

These deviations are in general superimposed and cannot be completely separated [38]. Especially in the micro range, inhomogeneities and temperature fluctuations lead to relatively large deviations compared to the tolerances. This has to be compensated for by quality inspection, because in general manufactured parts are designed and inspected at a reference temperature of 20 °C as described in DIN EN ISO 1.

All manufactured products show structural deviations from their desired micro- and macrostructure. According to the VDI/VDE-Guideline 2601, these two structural properties define the following deviations: The microstructure includes the roughness and waviness of the surface and the macrostructure covers size, form and position deviations. The corresponding size tolerances are defined for both length and angle deviations. In general, a size deviation can be positive or negative.

Form and position tolerances are defined in ISO 1101. Form deviations are determined based on the desired geometrical “shape” of a manufactured product. The nominal form must be mathematically describable, ranging from simple geometries such as circles, spheres or cylinders to complex geometries such as gear flanks, freeform surfaces or helixes. Based on these surfaces the corresponding tolerance zone is defined by a “generating” sphere or circle with a diameter equal to the form tolerance. By moving this generating element along the whole 2D or 3D nominal geometry, the surfaces limiting the tolerance zone are generated. In Fig. 10.2a an example for a roundness deviation based on a minimum zone/Tschebyscheff circle is shown. The tolerance zone defined by an annulus in Fig. 10.2b must contain all measured points for this feature.

Figure 10.2c shows the decision from a comparison of the actual deviation b_{actual} and the tolerated deviation $b_{specified}$. This is required for each feature of the manufactured product. If one feature is out of tolerance it has to be rejected. All tolerance zones for form deviations are either planar or spatial. ISO 1101 defines in detail the following 2D and 3D form tolerances:

- 2D: Straightness, roundness and line form
- 3D: Flatness, cylindricity and surface form

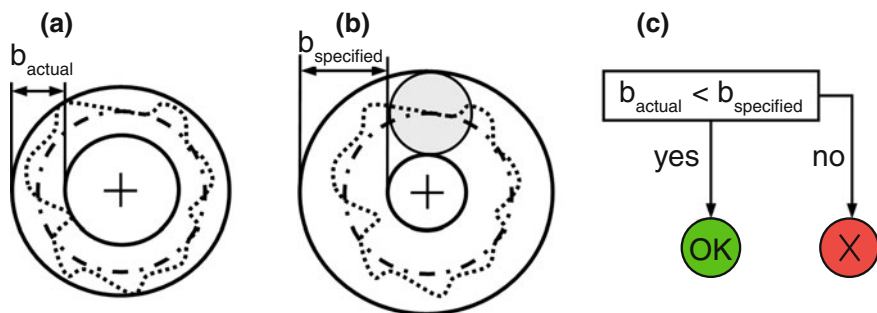


Fig. 10.2 Quality inspection of form feature roundness deviation: **a** evaluation of the actual deviation, **b** definition of the tolerance zone by a circle or a sphere and **c** decision based on comparison

Form deviations of spherical or conical surfaces are defined by surface form. A perfectly manufactured product would have a zero form deviation, thus a form deviation such as roundness deviation in Fig. 10.2 is always positive. Therefore, it is tolerated single-sided by an upper limit. The lower limit is per definition null.

The geometrical shape of a product is fully defined when all deviations are specified. Therefore, position deviations must also be tolerated. Position tolerances cover directions, positions and run-of features with respect to one or more datum elements. They are defined in ISO 1101, ISO 5458 and ISO 5459. The corresponding tolerance zone is based on datum elements, which have to be measured within the inspection of position deviations. The measured points of the tolerated geometry must lie inside the defined tolerance zone. Additionally, position tolerances may require nominal dimensions in the design drawing, which are highlighted with a rectangular box in the design drawings.

Not all features require a specific tolerance, because they may not be functional parts. Thus, these features can be tolerated by the general i.e. wider tolerances defined in DIN ISO 2768. Size, form and positions are generally tolerated with a remark in the design drawing. The lower character specifies the size tolerances and the upper character form and position tolerances. They vary on the specified size.

In general, all specified features of a product have to be measured. Due to modern machine tools and precise manufacturing most of the general tolerances can be quality inspected at a low sampling rate. All individually specified features should be inspected with appropriate measuring devices or gauges.

Modern measuring instruments are able to measure profiles with a high point density. This requires mathematical evaluations to calculate a geometric element e.g. a circle or a cylinder. Due to the definition of size, form and position tolerances up to four different evaluation methods can be applied. They lead to different results for the element, representing the totality of its measured points based on orthogonal distances d_{ai} (Fig. 10.3):

- a. Least squares element minimizing the square root of the sum of orthogonal distances d_{ai} between the individual n measured points with individual indices and the approximated ideal element, described by Eq. (10.1) [1, 18]. The calculated radius of the circle is r_{L2} .

$$L_2 - Norm : Q_G = \left(\sum_{i=1}^n [d_{ai}]^2 \right)^{\frac{1}{2}} \Rightarrow Min. \quad (10.1)$$

- b. Minimum zone or Tschebyscheff element describing an ideal element lying “in the middle” of all measured points (Fig. 10.2) and minimizing the zone enclosing all of the points with their distances d_{ai} , described by Eq. (10.2). The calculated radius of the circle is r_{sup} .

$$Supremum - /Tschebyscheff - Norm : Q_{sup} = \sup |d_{ai}| \Rightarrow Min. \quad (10.2)$$

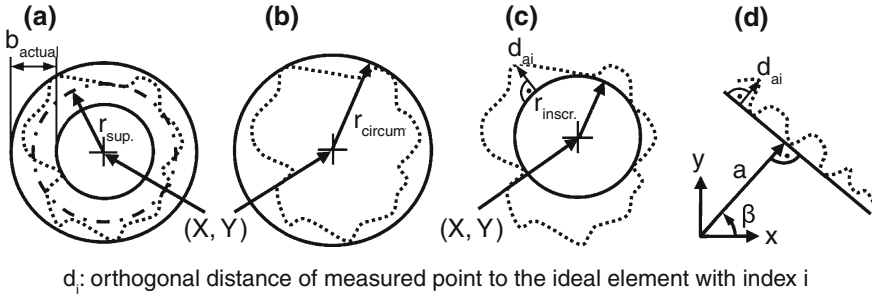


Fig. 10.3 Examples of evaluations in the quality inspection of manufactured parts: **a** evaluation of roundness deviation b_{actual} of a circle profile by a minimum zone/Tschebyscheff element (dashed circle), **b** minimum circumscribed circle, **c** maximum inscribed circle and **d** one-side fitted line

- c. Minimum circumscribed element enclosing all measured points with a minimum radius $r_{circumscribed}$, described by Eq. (10.3).

$$\text{minimum circumscribed elements : } r_{circumscribed} \Rightarrow \text{Min.} \quad (10.3)$$

- d. Maximum inscribed element with a maximized radius $r_{inscribed}$ such that all measured points are outside the calculated element, described by Eq. (10.4).

$$\text{maximum inscribed elements : } r_{inscribed} \Rightarrow \text{Max.} \quad (10.4)$$

- e. One-side fitted element, regarding e.g. lines and planes. These elements and also their evaluations are comparable to minimum circumscribed and maximum inscribed elements.

Figure 10.3 shows a circle profile evaluated as (a) minimum zone/Tschebyscheff, (b) minimum circumscribed and (c) maximum inscribed circle. A one-side fitted line is shown in Fig. 10.3d, where the line is described by a distance a and an angle β to the origin.

10.1.2 Tactile Metrology

Gert Goch and Karsten Lübke

For inspection, and to acquire the actual geometry and the corresponding features of a product, different tactile measuring and testing instruments are required. This comprises manual instruments like micrometer screws or calipers as well as numerically-controlled measuring devices in different designs. These tactile measuring instruments of different sizes have several positioning axes allowing

them to move a probing system with respect to the shape of a product and to obtain measuring points. Some examples of numerically controlled instruments are:

- Coordinate measuring machines (cmm) with or without a rotary table
- Form measuring devices
- Nano cmms
- Gear measuring instruments
- Measuring instrument for shafts, worms, screws, etc.
- Roughness measuring instruments
- Contour and roughness measuring instruments
- Various Scanning Probe Microscopes (SPM), e.g. Atomic Force Microscopes (AFM)

Tactile measurements imply that a probing force is applied to the measured object. These forces might be very low for micro parts, e.g. in the range of mN down to nN. But, the measured object has to be fixed during probing. Clamping is specifically adapted to the intended measuring task. It is required for access to the surface areas to be probed and for a reproducible alignment. Additionally, it can lead to the distortion of the part due to clamping forces.

Before the inspection process starts, a manufactured product is coarse and fine aligned within the measuring area of the device. Alignment describes a method to transfer the individual coordinate system of the specific product into the coordinate system of the design drawing. This transformation is based on the measured shape, showing deviations from the desired dimensions. It is performed by a mechanical or numerical transformation or a combination of both methods. Each measuring task may require a specialized measuring instrument. This leads to task specific stylus tip sizes and geometries. Common stylus and tip geometries to measure size, form and positional deviations and to measure surface roughness are shown in Fig. 10.4.

The stylus itself consists of a shaft and a tip. To avoid shaft probing in dimensional metrology, the size of the tip for size, form and positional measurements in general exceed the diameter of the shaft. The tips are manufactured from artificial ruby, ceramics or steel. Figure 10.4a and b shows:

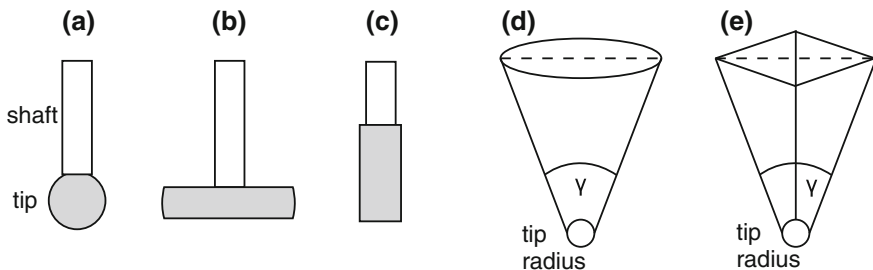


Fig. 10.4 Stylus tip geometries in tactile dimensional metrology: **a** sphere, **b** spherical disc, **c** cylinder, **d** cone and **e** pyramid with angle γ and tip radius

- a. Sphere: A spherical tip geometry is a common geometry in dimensional metrology. The sphere geometry is suitable for almost all kinds of surface. Its size ranges between 0.3 mm and several 10 mm with typical sphericity deviations of 50 μm or less.
- b. Spherical disk: The spherical disk is a cut sphere with a large radius, which leads to a mechanical filtering effect. It is applied e.g. for small ring grooves in boreholes.
- c. Cylinder: The cylindrical tip geometry, sometimes combined with a sphere tip, is applied for e.g. holes in flat sheet metal with small probing edges.

Surface measurements cover topography and roughness measuring tasks. The standardized tip geometries are defined according to ISO 3274:

- d. Cone form geometry.
- e. Pyramid form geometry.

In both cases, the tip of the cone is a very small sphere and common materials are diamond or steel. The standardized tip angle γ in Fig. 10.4 is 60 or 90° with a tolerance of $\pm 5^\circ$. Three different tip radii are defined: (2 ± 1) , (5 ± 2) and (10 ± 3) μm .

In tactile dimensional metrology different probing systems are common. They can be roughly divided into switching systems (with a touch trigger) and measuring systems (mostly scanning) sensitive in one, two or three directions. Generally, the stylus can be changed manually or automatically. All probing systems have to ensure their repeatability under varying influences: e.g. material, surface, temperature, vibrations, dynamics and dirt. Each stylus has to be corrected numerically according to its specific geometry. This geometry influences the measurement result [51].

To ensure a mechanical contact during probing contact, the stylus tip is pressed onto the surface with a certain force F by the probing system. This force is adjustable and has to be lowered for micro probes and parts. Therefore, every stylus acts as a cantilever beam, depending on its geometry, and thus bends. This bending is direction- and force-dependent and can be corrected numerically within the calibration routine of the measuring device. It affects the effective tip diameter and the shaft geometry. These influences are superimposed to the diameter and the sphericity deviations of the tip itself. Moreover, during the probing of the product surface flattening effects occur caused by the probing force. It applies Hertzian stress to the material (Fig. 10.5).

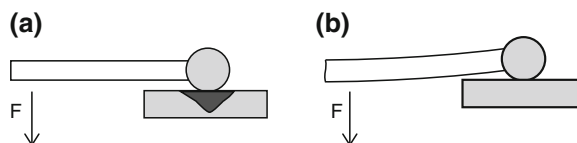


Fig. 10.5 Influences of stylus geometry: **a** Hertzian pressure and **b** bending applied by the contacting force F

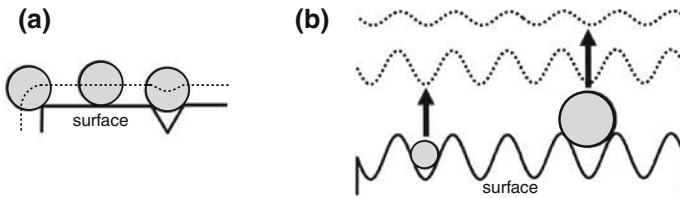


Fig. 10.6 Mechanical filtering by tactile probing: **a** smoothening of sharp edges and **b** lowered amplitude of periodical surface structures, depending on the probe tip diameter

All stylus tip geometries lead to an additional mechanical filtering. Therefore, the measured points, e.g. the coordinates of the probing ball centers, do not represent the real surface, but a filtered one. Mechanical filtering shown in Fig. 10.6 depends on the diameter of the stylus tip with respect to the local concave or convex curvature and curvature direction. This leads to very small tip diameters in roughness metrology to obtain a measured representation of the micro-topography [51].

In general a stylus has to be monitored with respect to possible changes in its tip diameter. Based on the number of measurements carried out with a certain stylus and depending on the surface material together with the surface topography, wear will occur on tactile stylus tips. This affects the correction of the probing ball, evaluating the “true” contact point between the surface and probing element. More details about tactile probing systems can be found in [51].

A completely different type of tactile quality inspection is realized by gauging (Fig. 10.7). Gauging in dimensional metrology describes the comparison of a geometric feature, e.g. a length or a diameter, with a reference standard. A double-sided tolerance zone is described by two gauges representing the lower and upper limit. To fulfill the double-side tolerated desired geometry one side must fit with the inspected feature and the other must not. The significant differences between measuring and gauging are:

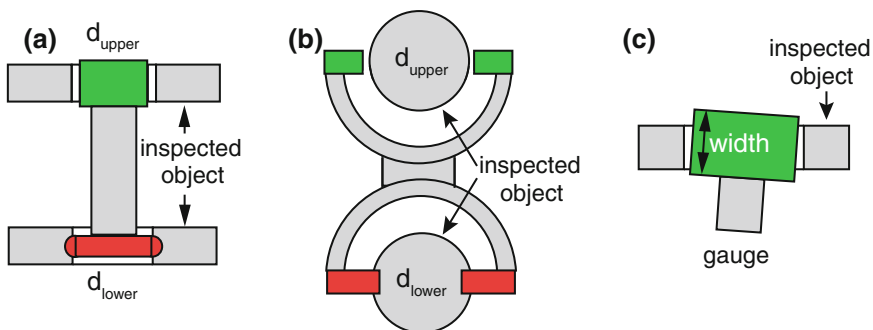


Fig. 10.7 Gauges for boreholes and shafts: **a** plug gauge, **b** limit roller gauge and **c** size relationship of “go”-gauge according to the Taylor-principle (length equal or larger than the depth of the inspected borehole)

- In contrast with measurement, gauging never leads to a measuring result, i.e. a quantitative amount multiplied by a unit.
- Gauging always includes an assessment of the inspected feature (“too small”, “too large”, “within tolerance”).

Both measurement and gauging can be combined for a complete quality inspection of a measuring object. In contrast to measuring instruments, gauges cannot determine a quantitative result. The result of gauging is “accepted”, “rejected” or “rework”, depending on the manufactured feature. In general they represent manual calipers, e.g. plugs or limit roller gauges for boreholes and shafts (Fig. 10.7). Many gauges have two sides, representing the upper and the lower limit of the tolerance zone. Often the “no go” side is marked red. According to the “Taylor-principle” plug ring gauges, for example, for the “go” side must fit completely into the borehole. Additionally, as shown in Fig. 10.7c, its cylindrical length should be larger than the borehole in order to inspect the whole feature. The “no go” side is engineered as a point-wise gauge contacting the borehole at only two points. With this design almost every local deviation out of tolerance can be detected. Many plug gauges do not follow this principle for several reasons, e.g. size, weight or cost. Both sides are e.g. designed as cylinders or as two-point calipers like the example in Fig. 10.7b.

With different types of gauge several inspections are possible. According to [38] this includes: size gauging, form gauging, position gauging, limit gauging and special gauging. Gauging is a very common process. In order to gauge micro features, the principle is limited by handling, clamping or accessibility issues.

Measurements in the micro range present specific challenges. When the styli get smaller and thus so do the tips, the measurements become more sensitive. Therefore influences that could be neglected in the macro range have an important impact in the micro range. Some of these influences are described in the VDI/VDE guideline series 2617:

- a. Dirt influences: In the micro range dirt can influence the measuring result. For tactile measurements the influence is lowered by the increased Hertzian pressure due to small probing balls and thus the dirt may be flattened. In general artifacts and measured workpieces should be cleaned accurately in every case to minimize these influences.
- b. Form deviations of the calibration artifacts: Measuring instruments and their styli are calibrated with artifacts. These artifacts are often spheres with low form deviations. Due to the high accuracy of coordinate measuring machines in the micro range these form deviations are relatively high in comparison with the diameter of the spheres, which influences the specified probing deviations.
- c. Calibration uncertainties of artifact features like sphere diameter: Coordinate measuring machines are monitored in repeated tests with calibrated artifacts. Due to the high accuracy of coordinate measuring machines in the micro range the geometric features of the artifact should offer lowered uncertainties based on their calibration. But such improved calibrations raise the required effort and cost.

10.1.3 Optical Metrology

Christoph von Kopylow and Ralf B. Bergmann

10.1.3.1 Description of Distinctive Features of Microformed Parts and Their Classification

Microparts have a size in two or more dimensions of less than 1 mm. For developing or optimizing a manufacturing process it is mandatory to precisely determine the shape or geometry of such microparts. Precise information on every step in a process chain is necessary to understand and to control the process. A deviation from the desired geometry always leads to restrictions and deviations in the process. Standard strategies, measurement systems and methods are usually not available for measurement tasks in the field of micro forming. For process support, 3D-coordinates with an accuracy down to 1 μm may be required, for example to control deviations in the shape or to get information about geometrical characteristics like radii or roundness. Very often the measured objects have a freeform shape without pass points, hence methods which measure area-wise are preferred. In contrast with tactile measurement systems, optical sensors work in a contactless manner and therefore do not affect the object under investigation, and are therefore especially suitable for micro components with sensitive and fine structured parts [3]. Optical sensors allow for both a higher working distance and a higher lateral resolution. Due to touch elements, the lateral resolution is about 300 μm for tactile systems and 15 μm for opto-tactile sensors. The optical approach allows for a (diffraction limited) lateral resolution down to the sub- μm area and a vertical resolution down to the nm range.

Micro parts have some properties which greatly affect optical metrology: high surface gradients or high surface angles. In addition the presence of both rough and reflecting parts of the surface changes the properties of the backscattered light which contains the information for the measuring system. The system should measure the complete or at least a large area of the objects' surface. However, certain areas of micro parts with high surface gradients will not scatter enough light back into the detector (camera). Hence the numerical aperture has to be increased which leads to a reduced depth of sharpness. In addition, the accessibility to the object is also frequently limited. For geometrical (non-interferometric) measuring techniques the observable surface angle can be increased by the illumination and measurement of the object from different directions. In [9] a method which varies the focus by vertical scanning of the probe and thereby providing topographical and color information is adapted by increasing the illuminating angle, which allows for measuring steep surface angles—unfortunately with a resolution of only several μm .

Figure 10.8 shows an example of a micro forming tool. Here, the abrasion in the marked area has to be detected with high resolution. Typically in micro cold

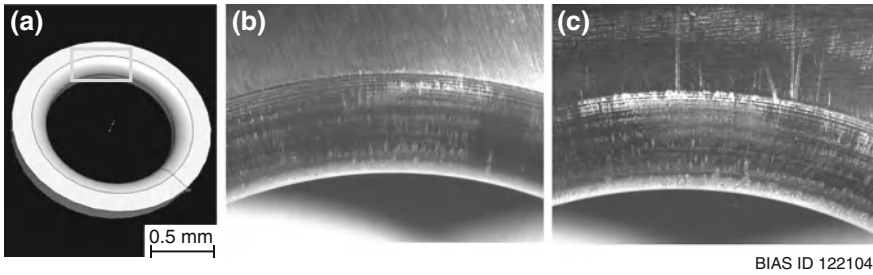


Fig. 10.8 **a** Micrograph of a deep drawing ring with a diameter of 1 mm, **b** cut-out of virgin tool, and **c** tool with deterioration

forming objects, such strong curvatures and high gradients are predominant. These attributes have a strong effect on the measurement process.

Figure 10.9a demonstrates the loss of light of a micro wire manufactured by micro rotary swaging, when measuring with an imaging system like a microscope. The wire with a diameter of 0.5 mm nearly fills the complete image, but only the central area scatters back enough light for measurement. Another effect is shown in Fig. 10.9b: the curvature leads to high fringe orders during an interferometrical measurement, which means the data cannot be resolved outside of the central region. In addition, the noise increases in the outer region as well due to the small amount of light in this area.

The surface of the object can be optically rough, which means that the roughness is as large or even larger than the measurement wavelength, or it can be specular, which means that the roughness is low compared to the wavelength. Optical measuring techniques usually cannot measure optically rough and specular surfaces at the same time, and the result of the measurement is affected by the roughness.

Figure 10.10a illustrates the resolution along the optical axis for the different methods of reflectometry, shearography and interferometry as a function of the roughness. It can be seen that reflectometry is restricted to smooth surfaces, whereas shearography can take measurements on rough but also partly on specular

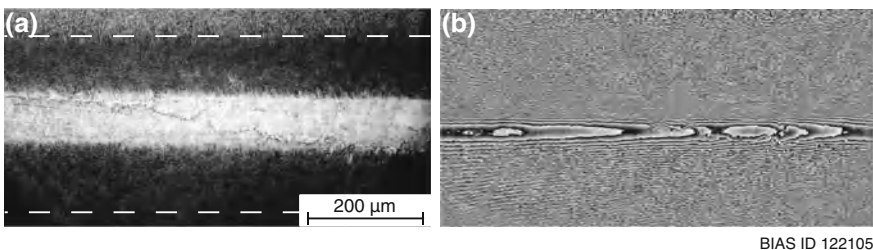


Fig. 10.9 Different measurements of a micro wire. **a** Camera image with reduced measurement area caused by the limited numerical aperture. **b** Interferometric phase with fast increasing fringe density caused by the high surface gradient

Table 10.1 Classification of objects which are optically observed

Property	Parameters of interest
Geometry	Border, roundness, chip formation, feathering, position, location, length, angle, flatness, freeform recording (3D-point cloud)
Surface structure	Structure recognition (e.g. as texture), roughness (tribology), reflectivity, defects, abrasion
Material	Change in material, color, layer (for example coatings or oxidation)

Table 10.2 Classification of boundary conditions which have an effect on the measurement process

Property	Parameters of interest
Macro-geometry	Macro-geometry with micro properties: handling relatively easy, tactile measurable (slow); example: drawing tools, (raw) material, tools, semi-finished goods Micro-geometries (with micro properties): problem stability (varying measurement results depending on location of measurement); example micro-cap, micro parts
Micro-geometry (measuring area): shape classification	Flat (panel sheet) Simple curved (edge): convex (micro-wire, cutting edge), concave (groove) Curved in two directions (muff): convex (cap from outside, deep drawing piston), concave (deep drawing ring) Freeform objects (part of a micro valve, micro structures with high aspect ratio)
Accessibility	Shadowing, maximum angles or gradients, undercuts
Handling	Stability, handling technique (fastener) and handling conditions

process. Table 10.2 shows as an example some boundary conditions which are based on the geometry of the object under investigation.

Table 10.3 gives an overview of well-known measurement techniques and their applicability to selected requirements of a measurement process of micro parts. As a consequence of the evaluation given in Table 10.3, micro fringe reflection and shearography appear to be particularly suitable for the metrology of micro parts, but one has to take into account that these techniques require an integration of the data to obtain the shape of the object. Hence they are prone to systematic errors. On the other hand techniques like holography directly deliver the shape and are applicable if the height range is small, what is often the case for micro objects due to their limited size.

10.1.3.2 Overview on Different Optical Methods for Micro Measurement

In the field of contactless measurement methods radiographic tomographic methods have to be taken into account. For measuring areas which are difficult to

Table 10.3 Schematic classification of different measurement techniques with respect to object properties

measuring method	contact-less (non-distort.)	robust	fast	suitable for specular surfaces	suitable for high surface gradient	capable of several μm height range	capable of sub- μm resolution
tactile coordinate measurement							
laser triangulation							
scanning confocal microsc.							
(micro) fringe reflection							
(micro) fringe projection							
interferometry							
white light interferometry							
holography							
shearography							

Legend: *dark* unsuitable, *light gray* suitable under specific conditions, *white* generally suitable

access or which are inside an object, micro x-ray computed tomography (μXCT) is the only applicable method for non-destructive measurement. The method reaches a resolution of some μm , in special cases down to the sub- μm range relative to a reference part [11]. It is necessary that the object is accessible from two sides and can be rotated. The technical effort is comparatively high and can only be applied when appropriate shielding of the ionizing radiation is possible. The resolution compared to optical measurement methods like white light interferometry is inferior; therefore once optical methods are applicable they should be preferred, which is the case for most micro metal parts.

Knowledge of the limits of the different optical techniques is essential for their proper use. For micro parts the main limits are, as mentioned above, given by a limited optical access and high surface gradients of the object under investigation leading to an insufficient quantity of light at the recording unit. In the following, suitable optical measurement techniques will be shortly described.

Up to now no universal measurement system has been available for micro measurements [26]. Therefore at present measurements are performed using various completely different sensors. The employment of a method depends strongly on the measurement task, the object and the boundary conditions. Very often the measurement problem can only be solved using multi-sensor equipment. An overview of methods which can be used in the micro range can be found in [17].

A classification of optical sensors can be done using the type of sensing: 1D (point sensor), 2D (line sensor) and 3D (area sensor) or considering the light source (laser/no-laser) or with regard to scanning or non-scanning methods. However, this does not reveal the physical behavior of the measurement method.

In addition, several techniques work point-wise and also area-wise. Hence a classification should be based on the physical principles:

1. Coherent Methods

These methods explicitly use the phase value of electromagnetic waves, therefore the use of temporal coherent light is mandatory. These methods are:

- Interferometry and digital holography [30, 41]
- Shear-interferometry [15, 45]
- Scanning methods:
 - White light interferometry (WLI) [14] with an endoscope [29] and with additionally adapted optics [2]
 - White light heterodyne interferometry (WHI): short coherent two wave-length interferometry [2]

2. Geometrical Methods

The following methods are based on the analysis of geometrical proportions and allow for the utilization of incoherent light sources. The methods are:

- Triangulation (for rough, scattering objects)
 - Focus-based triangulation: focus varying microscopy [9] and confocal microscopy
 - Scanning methods: line scanner and point scanner (laser scanner)
 - Photogrammetry (for example stereo photogrammetry)
 - Micro-fringe projection
- Micro-fringe reflection (for specular surfaces)
- Reflectometric stereo techniques for any freeform areas and miniaturized systems [25]

3. Measurement features and limits

The capability of a measurement method is characterized by the following attributes:

- Measurement area (size)
 - Lateral measurement area: inline methods (confocal sensor, WLI, focus variation) have combined with a low working distance (high numerical aperture) small possible measurement areas. For geometrical methods like fringe projection measurement areas with dimensions of several mm are typical [34]
 - Depth (deepness of the object)
- Resolution:
 - Best possible lateral resolution d is given by Abbe's resolution limit

$$d = 0.61 \frac{\lambda}{NA} \quad (10.5)$$

with $NA = n \sin \alpha$ with the refractive index $n = 1$ for measurement in air and the half angle of the aperture α leading to $d = 2 \dots 6 \mu\text{m}$ for $NA = 0.1 \dots 0.3$

- Depth resolution: dependent on the method correlated for example with the numerical aperture and may be as low as the nm range
- The physical limit of the minimum detectable feature size due to the uncertainty relation is discussed in [24]
- Dynamic range: height resolution compared to the measurement area. For many measurement systems it is about $1:10^4$; but for reflectometry it is for example $1:10^6$
- Accuracy and traceability

4. Boundary conditions of the measurement process

The choice and application of a measurement method also depends on boundary conditions where the most important are:

- Measurement time or velocity: here one has to balance between stability and accuracy (often determined by the production process!)
- Inspection costs: a rough comparison can be found in [2]
- Risk potential: for example the use of ionizing radiation for inspection
- Interaction of measurement method with the surface of the object [3, 17]
- Limitations for the measurable object geometry: surface curvature (for example fringe density in interferometry) and surface gradient (for example missing back-reflection)

In industrial applications, very often the inspection costs play a major role in the operation of measurement techniques [2]. In general measurement sensors for inspection in the micro range are usually considerably more expensive than corresponding methods in the macro range.

As an example, here we present measurements of a micro wire which have been also shown in Fig. 10.9. A good approach is using an area sensor. In Fig. 10.11 measurement of the micro wire with the help of a WLI (Fig. 10.11a) and a confocal microscope (Fig. 10.11b) are shown. The measurement area of the WLI is typically small (here $260 \times 260 \mu\text{m}^2$). The measurement is restricted to the top area of the wire, which makes it difficult to get precise information about the radius and roundness. But it allows for a detailed structural analysis of the surface properties. The confocal microscope also only delivers data from the top area (Fig. 10.11b, c) and is very noisy in the shoulders (turquoise curve in the line cut). The reason for this—the decreasing backscattered intensity—is displayed as green curve in Fig. 10.11c.

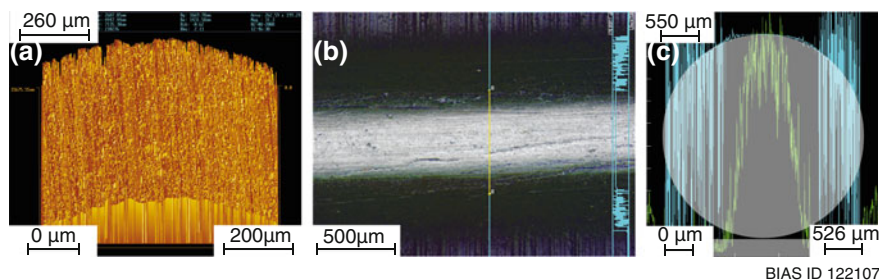


Fig. 10.11 Micro wire measurement: **a** using white light interferometry (measurement area $200 \times 260 \mu\text{m}^2$); **b** using confocal microscopy with a measurement area of $2 \times 1.5 \text{ mm}^2$; **c** line cut from **b**: *turquoise*: height, *green*: intensity; *bright transparent*: wire (measurement area $526 \times 550 \mu\text{m}^2$)

As for all optical methods, the measurement signal vanishes at steep edges in the micro fringe reflection technique. This can be seen from measurements of the deep drawing ring described in Fig. 10.8a.

Figure 10.12 describes measurement with the micro fringe reflection technique of a ring which is used as a micro deep drawing tool. The setup is shown in Fig. 10.12a. In Fig. 10.12b a line cut from the curved part of this ring can be seen illustrating the problem. Micro parts can be measured completely as long as the maximum angle of the surface is not exceeded, which is given for a light ray coming from the outer border of the reference plane via the surface into the aperture of the camera. For the fringe reflection technique the angle is not limited because of the optics from the capturing camera, but from the limited area on which the reflected fringes are generated. Hence a well-designed, size-adapted area of fringes can significantly increase the usable measurement area.

Table 10.4 shows an overview of several measurement techniques. The table shows the possible measurement volume, the reachable measurement precision and approximate costs of the measurement device. The costs largely depend on the degree of automation of the manufacturing process and the handling of the object.

10.1.3.3 Testing Strategy for Measuring a Micropart Using Digital Holograph

A testing strategy takes into account the optimal combination of object classification and qualified measurement method to solve the measurement problem, as well as the necessary process steps for the characterization and adaptation of the measurement method.

For measuring micro objects very often interferometric measurement techniques are particular suitable, because they measure area-wise, are contact free and have an accuracy of a fraction of the wavelength employed. In this context digital holography has been identified as an important tool [44]. In this method the object under investigation is illuminated with a laser and the backscattered light is superimposed with a reference wave. Digital holography allows for the

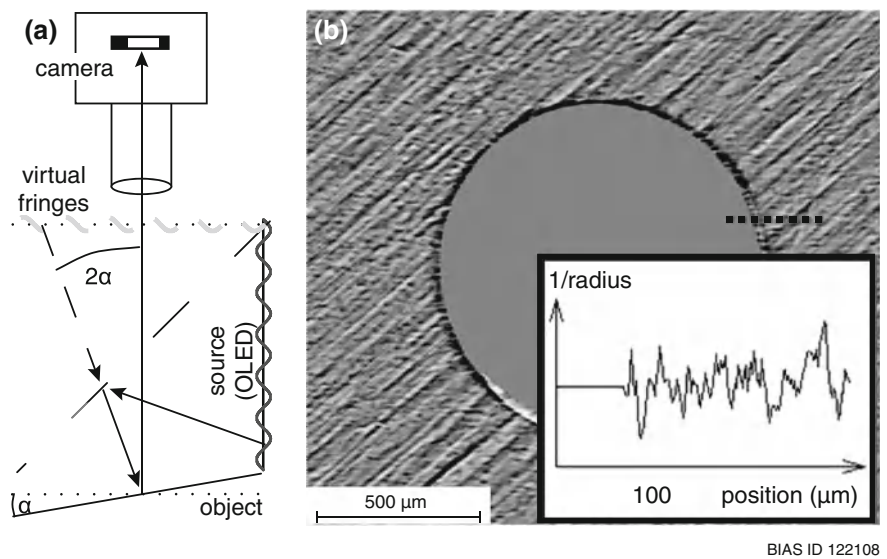


Fig. 10.12 Micro-fringe reflection: **a** schematic of a micro fringe reflection system. The fringe pattern, which is generated from an OLED, is reflected with the help of a beam splitter onto the object. For the camera it appears to be a reflex from a point in the virtual pattern plane with an angle of 2α to the light ray, which goes from the object to the corresponding camera pixel. **b** Micrograph of the measurement of a deep drawing *ring*. Inset along dashed line shows curvature data which can be measured very close to the edge of the *ring*

Table 10.4 Overview of different measurement methods including typical values for measurands, measurement volume, precision and costs

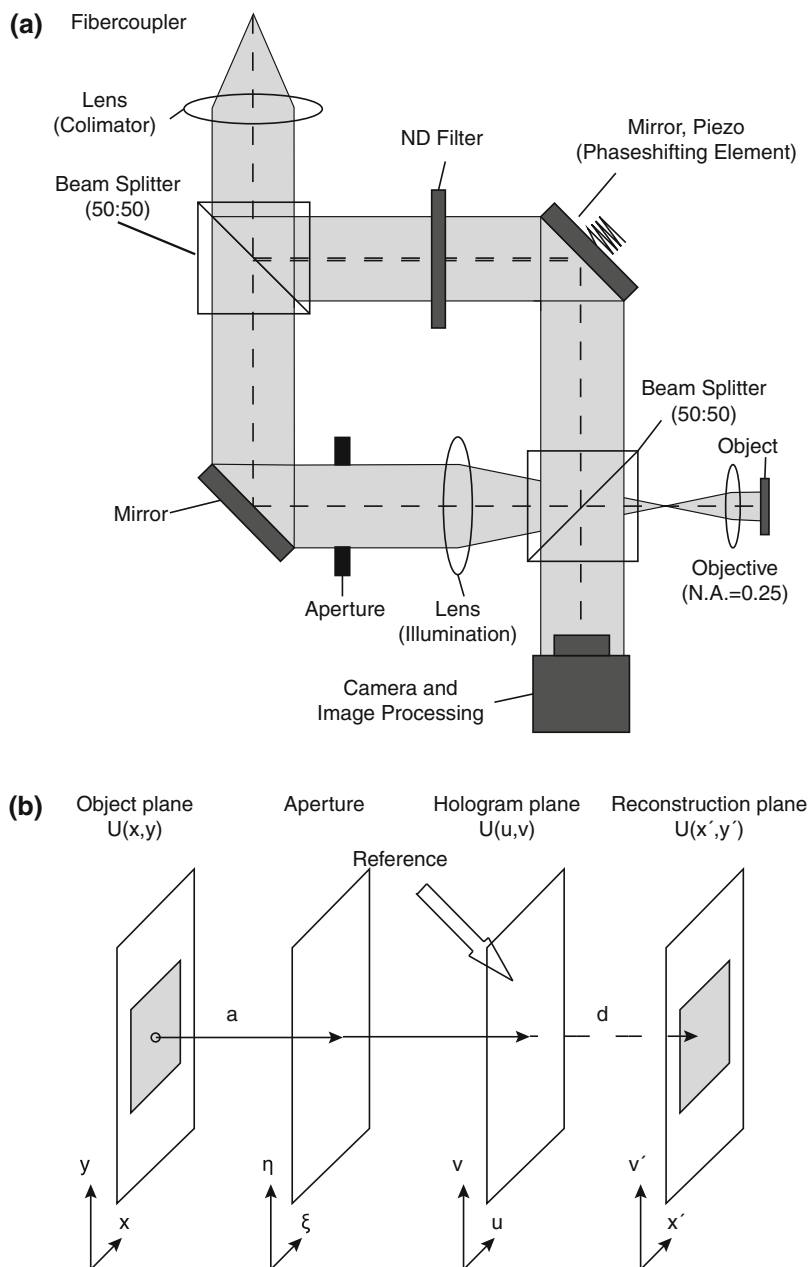
Measuring method	Measurand	Measured volume		Precision		Cost (order of k€)
		lateral (mm)	depth (μm)	lateral (μm)	depth (nm)	
Tact. coord.	3D coordinates	1,000	600 mm	2	2 μm	500
μ -fringe reflection	Surface angle $\Rightarrow$ 3D coordinates	5	500	40	1	50
μ -fringe projection	3D coordinates	20	5	40	1	50
Interferometry	3D coordinates or deformation	5	50	2	<1	50
Digital holography	Wave field (interference)	5	50	2	5	50
Shearography	Gradients $\Rightarrow$ 3D coordinates	5	500	2	25	50
Confocal microscopy	3D coordinates	1	7	130	20	100
Laser triangulation	3D coordinates	200	2	10	20	50
White light interferom	3D coordinates	1	1	5	<1	100

reconstruction of the complex wavefield at any depth in the computer, and hence the geometrical data of the object can be calculated. Usually the latter are determined from an analysis of the phase distribution in the reconstruction plane. To derive metrical data a geometry model of the interferometric setup including the optical path is mandatory. The properties of the objects' surface are essential. Typical microsystems very often have a reflecting surface which is made from aluminum, silica or copper. The size of the object is less than $1 \times 1 \text{ mm}^2$. The lateral resolution obtainable with digital holography is typically determined by the restriction of the numerical aperture due to the construction of the setup, as the necessary optical components such as illumination optics and beam splitter have to be placed and aligned between the object and sensor. This problem can be overcome using a microscope objective with high numerical aperture, which allows for the capture of the hologram in the image plane [52].

Figure 10.13a shows a sketch of a holography setup using a microscope objective. The collimated beam of a coherent light source is divided into a reference and an object wave by a beam splitter. The object wave encounters a lens via a reflecting mirror. The focal plane of this lens is located in the focal plane of the following microscope objective, resulting in an illumination of the object with parallel light. The second beam splitter superposes the light, which is backscattered from the object with the reference wave, and the camera captures the hologram.

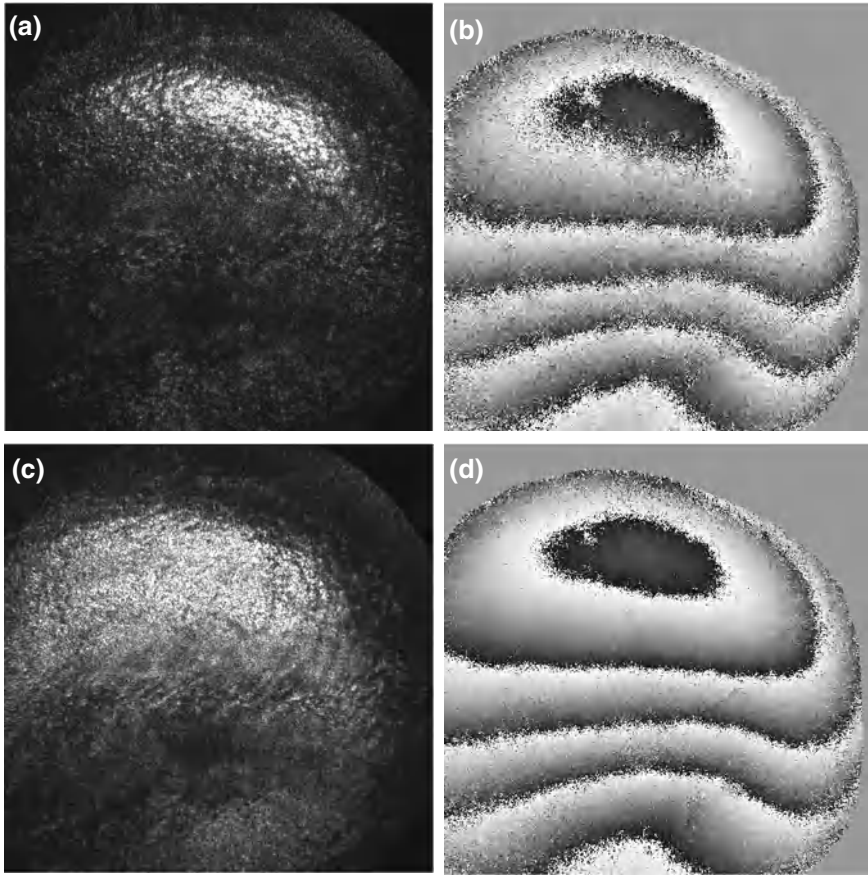
Figure 10.13b depicts the geometry model used for digital holography with a microscope objective: the complex field amplitude $U(x, y)$ which is backscattered from the object passes through the microscope objective with its aperture in the ξ - η -plane and propagates to the image area where it is superimposed with the reference wave, and the resulting hologram is captured by the camera in the u - v -plane. Using a phase shifter in the light path of the reference beam allows for applying phase-sampling techniques. For calculating the field distribution $U(u, v)$ in the hologram plane an algorithm is used, which requires four holograms. Each of these holograms has an increasing phase difference of $\pi/2$ with respect to the previous one. The field distribution $U(x', y')$ in the reconstruction plane is afterwards calculated by numerically solving the diffraction integral [28]. Because of the free choice of the reconstruction distance, digital holography is not bound to a fixed image plane. This is an advantage when measuring for example objects which are thermally loaded resulting in a thermal enlargement of the setup, which would move the object out of the focus plane in conventional microscopy. The shape of the object can be obtained from an analysis of the phase of the light in the reconstruction plane. If this is the image plane of the objective, the conditions correspond besides the magnification m to those directly above the objects' surface. Therefore the measurement result can be interpreted as a comparison of the surface topography with this virtual plane.

Figure 10.14 shows the results of measuring a microformed cup using the two-wavelength contouring method. The wavelengths used were $\lambda_1 = 580$ and $\lambda_2 = 583 \text{ nm}$, leading to a synthetic wavelength of $\Lambda = 112 \text{ }\mu\text{m}$. Figure 10.13a displays the reconstructed intensity distribution. The typical problem of measuring



BIAS ID 122109

Fig. 10.13 Experimental setup of a digital holographic microscope (DHM): **a** sketch of setup; **b** digital holography using a microscope objective (For explanation see text.)



BIAS ID 122110

Fig. 10.14 Determination of the shape of a micro cup. **a** Intensity and **b** phase from holographic reconstruction. **c** Intensity using three illumination sources. **d** Phase from holographic reconstruction with three illumination sources (length of the edge of all images $0.5 \times 0.5 \text{ mm}^2$)

such a micropart can clearly be recognized as shadowing effects due to the high surface gradients and decorrelation effects because of the surface structure. This leads to noise in the measured 3D shape data at points where not enough light is backscattered. This can be seen in Fig. 10.14b, where the phase data of the backscattered light which corresponds to the height data of the microcup are shown. Each line corresponds to a height difference of $\Lambda/2$. Hence the noise is in the range of several tens of micrometers depending on the illumination intensity. Figure 10.14c shows the intensity reconstruction when the object is illuminated from three directions. The corresponding height data in Fig. 10.14d have reduced noise. Using an algorithm which takes the illumination into account and weighting the data with the amplitude of the backscattered light, the three illumination

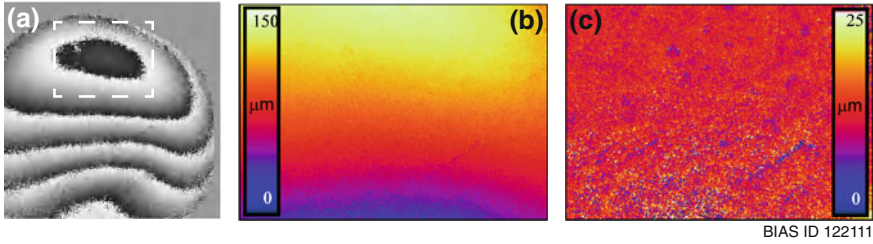


Fig. 10.15 Determination of the shape of a micro cup. **a** Phase from holographic reconstruction and marked cut-out area (length of the edge $0.5 \times 0.5 \text{ mm}^2$). **b** Calculated shape data. **c** Fine structure of object's surface after removing the shape of the object

directions lead to a better signal to noise ratio of 3 dB with a measurement uncertainty of less than 3 μm .

A decreased measurement uncertainty allows for a better analysis of the properties of the micropart. Figure 10.15b shows a cut-out of the measurement of the micro cup which displays the object's shape in this area. After removing the profile the fine structure of the object's surface becomes visible, which is displayed in Fig. 10.15c.

10.1.4 Measuring Uncertainty

Gert Goch and Karsten Lübke

According to the International Vocabulary of Metrology (VIM) or DIN 1319, the complete measurement result is generally expressed as a single measured quantity value together with the unit and a measurement uncertainty. An example is given in Eq. (10.6):

$$\text{Length} = (10,234 \pm 0,005) \text{ m} \quad (10.6)$$

The measurement uncertainty is a non-negative quantity. VIM defines it as a parameter characterizing the dispersion of the quantity values being attributed to a measurand. It further notes that the parameter may be a standard deviation or a specified multiple of it, called standard measurement uncertainty or the half-width of an interval, related to a stated coverage probability.

The measurement uncertainty always corresponds to a certain measuring process and not to a certain measuring device. A mathematical description of the measuring process is based on a model expressed e.g. by the function Y in Eq. (10.7), describing the physical or technical basis of the measurement.

$$Y = f(X_1, X_2, X_3, \dots, X_N) \quad (10.7)$$

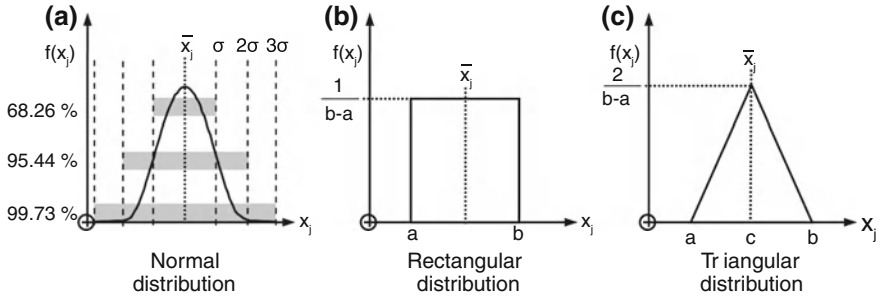


Fig. 10.16 Different distributions common for measuring uncertainty evaluations: **a** normal distribution and the respective coverage, **b** rectangular distribution and **c** triangular distribution

This measurement model comprises of several, in detail N , influencing input quantities X_i with individual standard measurement uncertainties u_i . These individual standard measurement uncertainties can refer to various distributions. Common distributions are e.g. normal, rectangular or triangular distributions (Fig. 10.16).

In order to calculate the measurement uncertainty for a specific measurement result of Y , comparable to Eq. (10.6), each of the influencing quantities X_i should have been measured several times such that one of the distributions in Fig. 10.16 can be applied reasonably. I.e. for each of the quantities X_i a limited number of measured data $x_{i1}, x_{i2}, \dots, x_{in}$ are available. They result in a set of N mean values $\bar{x}_i$ and N empirical standard deviations $\sigma_{emp, i}$, one pair for each influencing quantity X_i (see Table 10.5 and Fig. 10.16). Since the set of measured values has to be limited, $\bar{x}_i$ and $\sigma_{emp, i}$ can be regarded as (best possible) estimations of the corresponding expectation values ε_i and standard deviations σ_i of the influencing quantities X_i .

As a result, the standard measurement uncertainty of a measured or evaluated value y_i of the measurand Y is based on these individual uncertainties and thus called the combined standard measurement uncertainty.

The method behind this combination of random uncertainties is the theory of errors and error propagation by Gauß [19]. The method presented here does not cover unknown systematic errors, which are a challenge in metrology. Gauß [19] included systematic errors in his model about 200 years ago. They were introduced in international standards in around the millennium. The combined standard measurement uncertainty for the expectation value y of Y according to the model of Eq. (10.7) is given in Eq. (10.8). It is valid for non-correlated influences.

$$\sigma_{emp,y} = \sqrt{\sum_{i=1}^N \left(\left(\frac{\partial f}{\partial X_i} \right)^2 \cdot \sigma_{emp,i}^2 \right)} \quad (10.8)$$

Table 10.5 Expectation value based on limited measurement series and corresponding empirical standard deviation of selected distributions

Distribution	Expectation value based on limited measurement series	Empirical standard deviations
Normal distribution	$\bar{x}_i = \frac{1}{n} \sum_{j=1}^n x_j$	$\sigma_{emp,i} = \sqrt{\frac{1}{n-1} \sum_{j=1}^n (x_j - \bar{x}_i)^2}$
Rectangular distribution	$\bar{x}_i = \frac{a+b}{2}$	$\sigma_{emp,i} = \frac{b-a}{2\sqrt{3}}$
Triangular distribution	$\bar{x}_i = \frac{a+b+c}{3}$	$\sigma_{emp,i} = \frac{1}{6} \sqrt{(a-b)^2 + (b-c)^2 + (a-c)^2}$

Table 10.6 Factor k and the corresponding probability Pr of a given percentage of values is within the interval

Factor k	Probability Pr in %
0.676	50
1	68.27
1.645	90
1.96	95
2	95.45
2.576	99
3	99.73

Nonetheless, most of the individual uncertainties u_i can be evaluated according to the Guide to the Expression of Uncertainty in Measurement (GUM) published by ISO/BIPM [5]. It defines two evaluation methods:

- a. Type A: evaluation of the measurement uncertainty from a series of measured values obtained for the quantity and its characterization by standard deviations (referring to an assumed statistical distribution).
- b. Type B: characterization by standard deviations, evaluated from probability density functions based on experience or other information.

To cover a certain probability with the combined standard measurement uncertainty, $u(x_1, \dots, x_N)$ can be multiplied with a factor k larger than one. This results in the expanded measurement uncertainty $U(x_1, \dots, x_N)$ in Eq. (10.9). The VIM [6] notes that the factor depends upon the type of probability distribution of the output quantity in a measurement model and on the selected coverage probability.

$$U(x_1, \dots, x_N) = k \cdot u(x_1, \dots, x_N)$$

(10.9)

If the combined standard measurement uncertainty u is normally distributed, Table 10.6 shows the coverage probability. The more the individual uncertainties are combined, the more the resulting distribution approaches a normal distribution and thus the factor k can be applied.

10.1.5 Quality-Oriented Process Planning

Gert Goch and Karsten Lübke

In order to assess a result like the one in Eq. (10.6) together with its expanded measurement uncertainty, it has to be compared with its nominal value and the corresponding tolerances. They are specified in the design drawing and can be double- or single-sided. Figure 10.17 explains the principle for two examples. Features with a double-sided tolerance, e.g. a diameter of a borehole, offer an upper specification limit *UL* and a lower specification limit *LL*. When the measurement result of the feature holds these limits, the feature is accepted. Single-sided features, e.g. form deviations, are per definition non-negative. They are tolerated with an upper specification limit, which should not be exceeded by the measurement result.

However, all measurement results, namely every feature measured within the quality inspection of products and their components, should be precisely confined by the corresponding measurement uncertainty. Thus the measurement specifies an interval, within which the correct quantity value can be found with a specified probability. In Fig. 10.17 both examples (a) and (b) show three different cases:

- (a) The uncertainty interval of the measurement result is completely inside the tolerance → the measured feature is OK
- (b) The uncertainty interval of the measurement result is completely outside the tolerance → the measured feature is not OK
- (c) The uncertainty interval of the measurement result overlaps with one limit of the tolerance zone → the quality assessment of the proposed measured feature is ambiguous

To avoid case (c), in 1968 Berndt et al. proposed the so-called “Golden Rule of Production Metrology” [4]. It states that the measuring uncertainty should be less than one tenth of the tolerance. On the one hand, even today it is not always possible to fulfill this requirement with modern measuring instruments, e.g. in gear

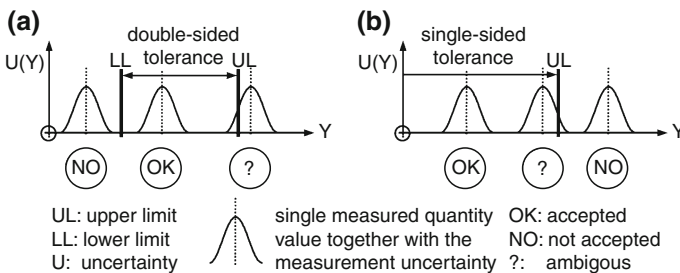


Fig. 10.17 Interaction between tolerance zone, measuring uncertainty and quality assessment: **a** double-side tolerated feature, e.g. length and **b** single-side tolerated feature with upper limit, e.g. form deviation

metrology [21]. On the other hand, the tolerances in production metrology are more and more reduced. Thus the interaction between tolerances, measuring uncertainty and quality assessment must be taken into account when measuring product features. Therefore, ISO 14253 defines decision rules for proving conformance or non-conformance with specifications [35]. They differ between manufacturers and customers:

- When a manufacturer inspects a product and measures a geometric feature, he must prove conformance with the tolerance zone, uniformly reduced by the expanded uncertainty U . The measured result must hold the limits of this reduced tolerance zone to accept it.
- When a customer inspects a product and measures a geometric feature, he must prove conformance with the tolerance zone, uniformly expanded by the expanded uncertainty U . The measured result must hold the limits of this expanded tolerance zone to accept it.

Thus, an influence of the expanded measuring uncertainty on the statistical control of a process exists. Manufacturing includes the generation of several features and results for one produced part. Each of these individual features is related to its corresponding manufacturing process, including one or more machines. For a regular production process it is assumed that all individual processes are in a state of statistical control and show a stationary and regular process behavior. Unfortunately, most processes tend to change, because of e.g. tool wear, and do not remain in a stationary and normally distributed state. ISO 21747 provides methods for evaluating the quality capability and performance of manufacturing processes with respect to time. They are categorized by the stability of the first and second distributional moments. This includes tests as to whether these criteria are constant, change systematically or randomly.

ISO 21747 requires extensive process analysis and subsequent process stabilization or improvement measures over time. They enable the efficient and effective control of a process, resulting in a higher quality of products. For the process analysis samples are taken from the process flow. The number of samples and the frequency of their analysis depend on the process type and the type of product. All important changes should be detected in time. The measurement results obtained from these samples lead to time-dependent functions, described by the distribution, the location, the dispersion and the shape of the results obtained from the analyzed samples. These time-dependent distributions can be classified into four main distribution models A, B, C, D with several sub-models, all based on the location and dispersion moments of individual (instantaneous) and resulting (outcoming) distributions. Figure 10.18 shows three examples of such time-dependent distribution models.

Based on the characterization of an individual product feature, performance and capability indices can be determined. The location and dispersion characteristic is calculated with respect to the tolerance of the feature. For this purpose, four methods M1–M4 are defined according to [36]:

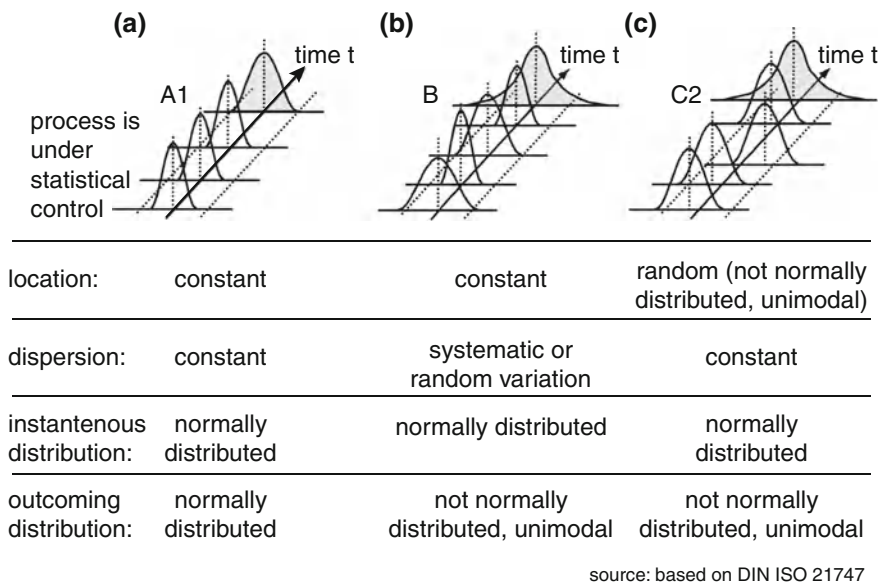
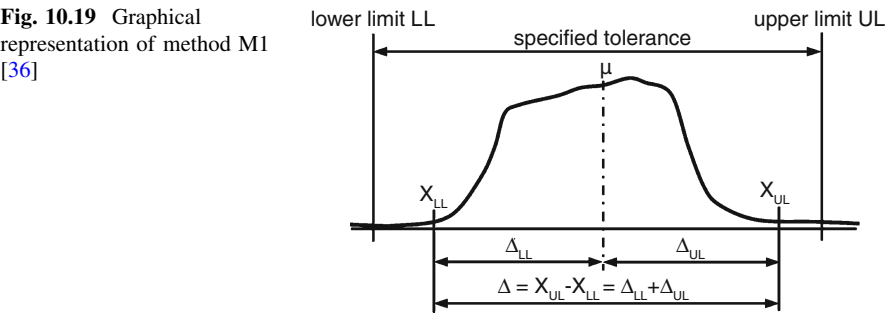


Fig. 10.18 Time-dependent distribution models for characterizing processes, defined in DIN ISO 21747: **a** distribution model A1, **b** distribution model B and **c** distribution model C2

- a. Method M1: General geometric method. Calculations with estimators for location μ and dispersion Δ .
- b. Method M2: Explicit inclusion of additional variation. Calculations with estimators for location μ , dispersion Δ and additional fluctuations μ_{add} .
- c. Method M3: Alternative method of explicit inclusion of additional variation. Calculations with estimators for location μ , dispersion Δ and additional fluctuations μ_{add} .
- d. Method M4: Excess proportions approach. Calculations with estimators for upper and lower excess proportions.

Process performance indices are defined by ratios of a length parameter of the distribution to the specified tolerance [36]. For the general geometric method M1 μ



indicates the location of the process and Δ the dispersion of the process, shown in Fig. 10.19. The dispersion Δ is bound by the lower quantile X_{LL} and the upper quantile X_{UL} .

The corresponding process performance indices are defined by Eqs. (10.10)–(10.13)

$$\text{process performance index } P_p = \frac{UL - LL}{\Delta} \quad (10.10)$$

$$\text{lower process performance index } P_{pkLL} = \frac{\mu - LL}{\Delta_{LL}} \quad (10.11)$$

$$\text{upper process performance index } P_{pkUL} = \frac{UL - \mu}{\Delta_{UL}} \quad (10.12)$$

$$\text{minimum process performance index } P_{pk} = \min(P_{pkLL}, P_{pkUL}) \quad (10.13)$$

When a process is in a state of statistical control its capability can be described by a process capability index. The corresponding formulae are defined by Eqs. (10.14)–(10.17), and are the same as for the corresponding performance indices:

$$\text{capability index } C_p = \frac{UL - LL}{\Delta} \quad (10.14)$$

$$\text{lower capability index } C_{pkLL} = \frac{\mu - LL}{\Delta_{LL}} \quad (10.15)$$

$$\text{upper capability index } C_{pkUL} = \frac{UL - \mu}{\Delta_{UL}} \quad (10.16)$$

$$\text{minimum capability index } C_{pk} = \min(C_{pkLL}, C_{pkUL}) \quad (10.17)$$

DIN ISO 21747 defines different estimators for the location μ and the dispersion Δ of a given data set [36]. It notes that no quantitative comparison of the indices is feasible, because the indices are calculated according to different methods. To keep a process in statistical control continuous monitoring is required. The standard solution is the use of different control charts. Today they are implemented as software tools allowing an easy backup of data and evaluation [31].

The process capability indices C_p and C_{pk} should be at least greater than or equal to 1. When C_p is equal to C_{pk} the expectation value lies in the middle of the tolerance zone. With a decreasing index C_{pk} the expectation value moves away from the middle. A common requirement in industry is both indices greater than or equal to 1.33. While $C_{pk} = 1.0$ corresponds to a minimum distance of 3σ to the next tolerance limit, $C_{pk} = 1.33$ corresponds to 4σ with around 66 failure parts per

million. According to the “Six Sigma” philosophy $C_{pk} = 2.0$ is required more frequently [32] and leads to less than one failure parts per million. For values of $C_p/C_{pk} \geq 1.33$ the production process is considered to be well controlled.

10.2 Closed-Loop Control

Gert Goch, Peiran Zhang and Volker Renken

In general, control engineering aims to manipulate a technical variable to achieve a desired value. So the error between the controlled and desired values should be zero. If the system does not have any disturbances this aim could be achieved by an open-loop-controller, where the manipulated variable of the plant is only affected by the desired value. However, most technical systems experience disturbances, which affect the plant in an unknown way. Therefore these systems demand a closed-loop-controller with a feedback of the measured controlled variable as the input for the controller. The measured controlled variable is compared with the desired value in order to eliminate their difference [13]. In the field of traditional control engineering, linear system theory became the basis for analyzing the control system. Therefore conventional controlled variables are usually the directly measurable process variables, such as the actual position of a tool or the rotational speed of a motor.

In a quality control system, it is essential to directly control the quality features of a workpiece (instead of the process variables creating these features), such as size (e.g. diameter of a bearing ring) or form parameters (e.g. edge radius), which are affected by both process variables and unplanned disturbances. According to the DIN ISO 8402 standard, quality can be defined as follows [39]:

Definition Quality is the entirety of characteristics belonging to a unit, with respect to their ability to fulfill defined or preconditioned requirements.

In general, the objective of a control system is to ensure certain quantities—in engineering applications usually output quantities of technical processes—at a specified target value. Besides the desired quality features, the quality requirements (e.g. tolerance) also represent important input information for the quality control system and must be defined before the production phase. In principle, a quality control system consists of an input block to define the quality features and quality requirements desired, which are compared with the actual measured values of the quality features. The metrology in Sect. 10.1 could be applied for the measurement of quality features. The controller applies a quality control strategy to influence the manipulated variables, which, together with possible disturbances, represent the input of the controlled process (Fig. 10.20).

Since the relationships between quality features and manipulated variables are usually complex and nonlinear, classical linear controllers (e.g. PID-controllers or Riccati-controllers) are not suitable to enable quality control. Therefore in the field

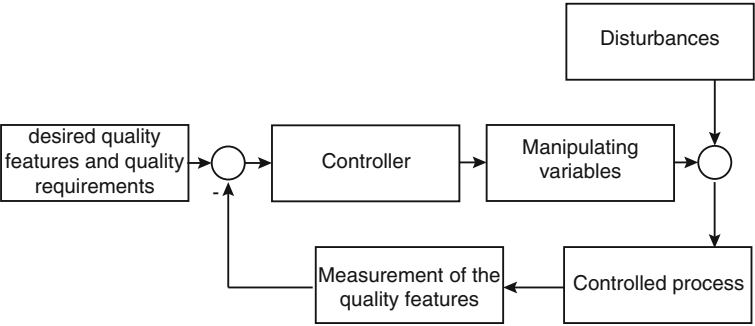


Fig. 10.20 Quality control system

of quality management various methods are applied, e.g. SPC (Statistical Process Control) and quality control charts that are used post-process. For in-process quality control, process models and measuring methods are needed.

10.2.1 Types of Control-Loops According to the Measuring Methods

In Sect. 10.1 different measurement methods for microstructures were introduced. Within control engineering, the applied measuring techniques can be divided into three groups: post-process, in situ and in-process. Furthermore, the control loop as well as the feedback can be realized on three levels: in-process, near-process and at cross-process level. The in-process level controls the quality features or the process variables for each subprocess. Here a real-time model is applied to operate the machine by manipulating variables. Based on a subprocess model and on in situ measurement, the near-process level controls the quality feature for the subprocess. The cross-process level controls the desired quality features and requirements

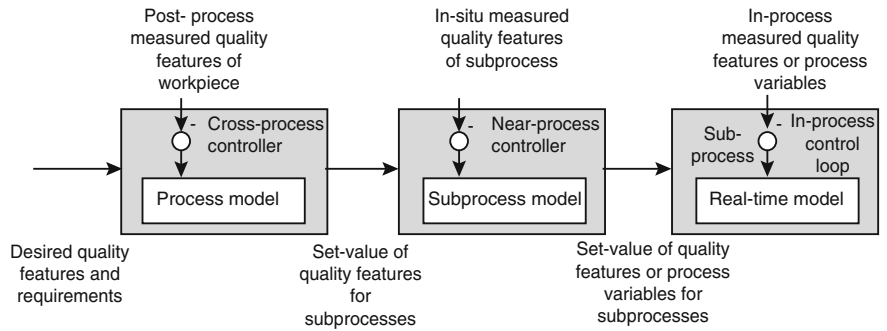
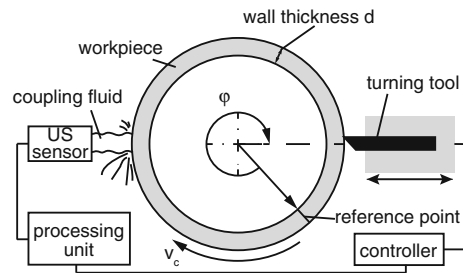


Fig. 10.21 Levels of a quality control system

Fig. 10.22 Closed-loop quality control of wall thickness in the turning process of bearing rings, based on in-process measurement



of the whole process. Therefore a whole process model is necessary, which could be built of several subprocess models. The final processed structures should be measured and used to optimize the manufacturing activities in the future (Fig. 10.21).

At the in-process level, the controlled variables are measured and fed back while the subprocess is in action, in order to adjust the processing in real-time. A classical linear time-invariant control loop for the process variables is assigned to this level. A closed-loop quality control system based on in-process measurement was a new development in recent years. Figure 10.22 shows an example in the macro range. It was designed for the turning of bearing rings using a cascade control concept. The quality feature of the product (wall thickness of the bearing ring) is controlled by in-process ultrasonic measurements [48]. The in-process quality control system for micro products should be designed with similar elements: controller, process model and in-process quality measurement.

Near- and cross-process control is applied at the upper (2nd and 3rd) levels of quality control systems. In near-process control loops, the quality features are identified when the manufacturing process is interrupted or when the product has left the actual subprocess. The applied measurement is used to optimize the subprocess and to recognize improved operating procedures and process values. As the near-process control is time discrete during production, in situ (workpiece still in the production facility) or near-process (outside machine tool) measurements are applied. Statistical process control (SPC) is one reasonable method of near-process control, which is mainly applied in mass production, e.g. for the electroplating of heated rear car windows [50]. By using a quantitative process model, the correlation between the variation of a quality feature and the related process variables can be explicitly described. Within the great variety of mathematical modeling approaches, physical equations are the classical method to describe a process. An approximation, e.g. by polynoms, is used to represent a nonlinear system. A practical manufacturing process dependent on multiple process variables usually shows complex interactions, where the cause-effect relationships are difficult to describe. In this case modeling by an artificial neural network is a common solution. The neural network is based on a large number of simple units. Conventionally it is robust against errors of single experimental data. Such a model can be optimized at all times with new data [33]. Hence, it can self-adapt to small changes of the process (e.g. changes in environmental conditions).

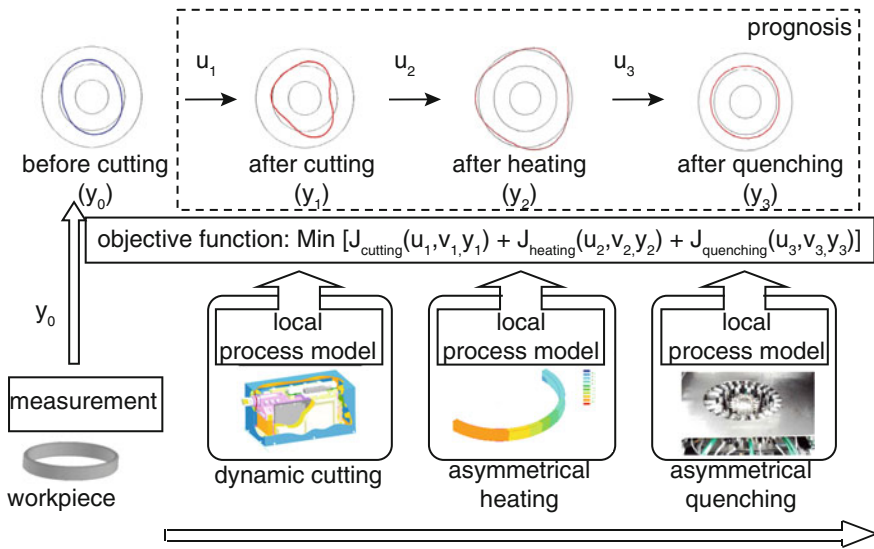


Fig. 10.23 Cross-process quality control concept for the production of bearing rings [12] J objective function, y quality features (roundness), u process variables, v required effort (production time, required energy, etc.)

Cho et al. [8] presents a near-process quality control based on a Polynomial Neural Network (PNN). The model is trained with data obtained from the near-process measurements of finished parts.

If the quality features of a workpiece are affected by more than a single subprocess, optimization throughout the whole process chain is necessary. Here the cross-process control links various subprocesses of a process chain and determines the optimal configuration for the production process. Therefore, the cross-process control also includes a feed-forward of information. For example, it determines the desired quality features for the remaining subprocesses.

In order to explain the principle of cross-process control, an approach realized for the production of bearing rings [22] is illustrated in Fig. 10.23. This cross-process control is based on the individual subprocess models. Having finished a certain subprocess, the cross-process controller transfers the remaining quality feature deviations of the workpiece to the next subprocess model, which predicts the quality feature values according to its actual input process variables. At the end of the process chain the resulting quality feature deviations are calculated. An objective function is used to quantitatively describe the performance of all subprocesses. Through minimizing the objective function, the optimal set values of the quality features are determined for the current and subsequent processes.

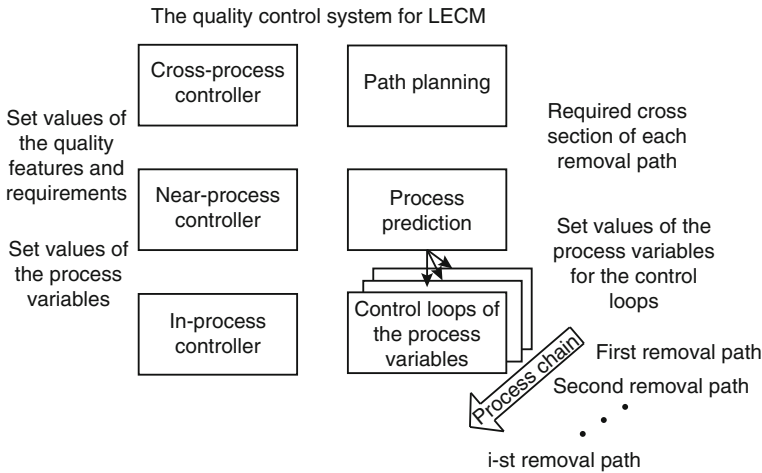


Fig. 10.24 Quality control system for laser induced electrochemical machining (LECM)

10.2.2 Application of Quality Control Systems in Micro Manufacturing

An exemplary quality control system in three levels is implemented in the field of micro manufacturing, specifically for *Laser-Induced Electrochemical Machining* (LECM, q.v. Sect. 7.3.3). For this process, the objective quality feature is the defined form deviation of the manufactured workpiece contour. In a preliminary study, the control system of the LECM process concerns three process variables, which affect the quality features (i.e. the cross-section of the removed material): laser power, feed rate of workpiece and flow rate of the etchant. The desired contour is achieved by a sequence of overlapping removal paths. Each removal path can be considered as a subprocess, where its cross-section deviations are the quality features for this subprocess (Fig. 10.24).

For the LECM system the cross-process level calculates the quality variables for the subprocess, and the set-values of the process variables for the in-process controller are determined at the near-process level. The near-process controller is realized by an inverse process model based on an artificial neural network and an iterative optimization algorithm. By using this near-process controller, the desired

process variables for sub-processes can be calculated according to the required cross-section of each removal path.

The cross-process controller links the sub-processes, the individual removal paths, and designs its positions as well as its cross-sections to achieve the desired smooth contour. The cross-process controller predicts the form accuracy of the workpiece. The deviations between the predicted and the post-process measured contour profiles are recorded and saved in a database. Hence, the process accuracy constantly improves [54].

10.3 Quality-Oriented Design of Process Chains

**Bernd Scholz-Reiter, M. Lütjen, D. Weimer,
D. Rippel and H. Thamer**

10.3.1 Process Chain Design in Micro-Manufacturing

The design of efficient micro production process chains is one of the most challenging tasks in the product development process. Before a new product can be launched, the production process has to be designed to achieve the given volume and quality requirements. Thereby, a lot of interdependencies along these multi-stage manufacturing process chains have to be considered. Due to the size effects, there is a lack of stable processes and standardized interfaces for the manufacturing, handling and quality inspection technologies in micro production. Technical requirements and capabilities of technologies vary greatly, while tolerances become increasingly small with the decreasing size of the products. For instance, the result of a deep drawing process can vary when using the same metal sheet because of the different orientations of the grains. Therefore, the design of micro process chains requires the consideration of the specific characteristics of the product, as well as of those technologies applied in manufacturing. A quality-oriented design method helps identify process sequences, which best meet the requirements. The idea of quality-oriented design in micro-manufacturing is to itemize complex technical as well as logistical correlations along the process chain and relate them to the manufacturing system, which is used to execute the process chain.

A micro manufacturing system is characterized by processes with high production rates of up to 450 parts per minute, with a throughput of several hundred to hundred-thousand parts per series. In addition, production tolerances with geometrical deviations below one micrometer have to be satisfied. In order to achieve such high production rates at a satisfactory product quality, micro process chains have to be designed very carefully, considering the interdependencies between different processes and technologies. Thereby, quality control and handling operations are integral to the production process and have to be adapted to the

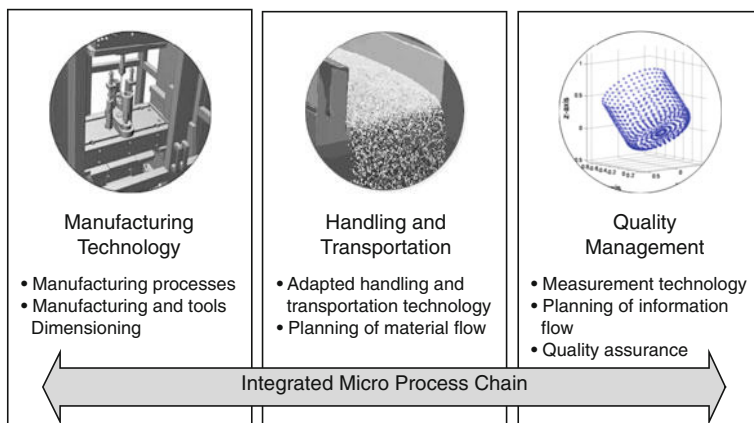


Fig. 10.25 Integration of different process types in micro manufacturing process chains

speed and tolerances of the manufacturing process. Consequently, an integrated planning of process chains has to consider the manufacturing, handling and quality inspection processes (Fig. 10.25).

By modeling such dependencies as a part of the process chain, more robust process chains can be designed, and issues between different processes or technologies can be identified. Ideally, this will be done in the early design phases and parallel to product design. This chapter introduces the general concept of process chains. Thereby, it discusses different techniques for designing and evaluating process chains. Afterwards, it presents a procedure for the quality-oriented design of process chains for micro production processes. Finally it demonstrates the integration of a quality control process and a micro deep drawing manufacturing process.

10.3.2 Fundamentals of Process Chain Design

Process chain design is one of the most important factors for efficient production. It can be done after or parallel to product design. By using the concept of simultaneous engineering, process design is integrated into to the product design and will be conducted parallel to it [16]. The objective of simultaneous engineering is to cease the sequential work flow of product and process design. The idea is to prepone work packages from process design and conduct them in parallel with product design. Therefore, preliminary planning results are used for the coordination of product and process design. This leads to a high demand on communication, because of the many small changes which have to be done during the design process. The idea is to avoid cost-intensive large changes in later phases. In general, the design of a process chain has to deal with the same

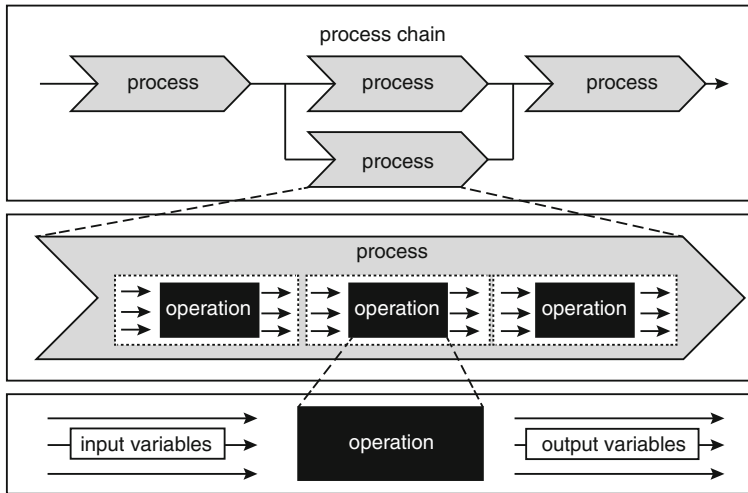


Fig. 10.26 Visualization of process chains, consisting of processes and operations

problems as other engineering design tasks. Usually the design process is faced with a very ill-structured problem. This problem has to be solved by achieving an objective with limited time and resources. *Tong* defines the term “design” as followed [49]:

What is design? Design is the process of constructing a description of an artifact that satisfies a (possibly informal) functional specification, meets certain performance criteria and resource limitations, is realizable in a given target technology, and satisfies criteria such as simplicity, testability, manufacturability, reusability etc.; the design process itself may also be subject to certain restrictions such as time, human power, cost, etc.

Based on this definition, the design of process chains is about the creation of a temporal and logical sequence of processes, which is able to transform given inputs into defined outputs. Thereby, a substantial distinction between the terms “operation”, “process” and “process chain” is necessary. The following definition provides three different levels (Fig. 10.26).

The first level is about the “operation”, which represents the smallest unit of a process chain. Each operation includes input and output variables and refers to an activity, for example the drilling of a hole [10]. The second level is about the “process”, which consists of several operations. Such an operation sequence is always linear and has to be assigned to at least one machine or resource. The combination of several processes into one process chain takes place at the third level.

In order to design high performance process chains, it is necessary to use knowledge from former implementations. In particular, the phases of configuration and evaluation need such information in order to estimate the behavior of the process chain in regard to the technical feasibility, cost-efficiency and product quality (Fig. 10.27).

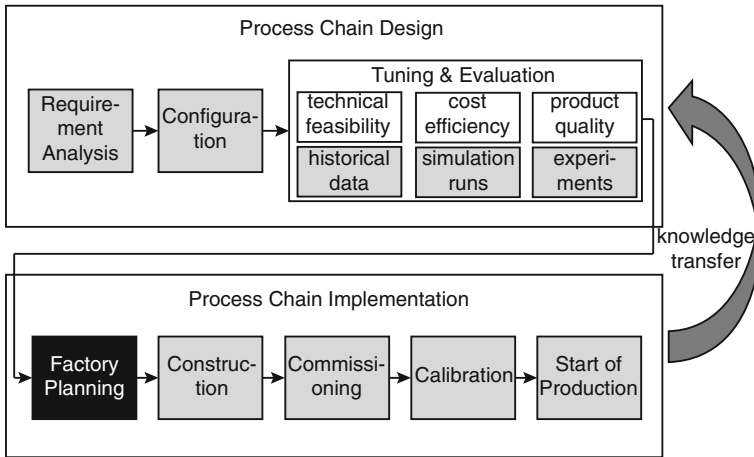


Fig. 10.27 Procedure model of process chain design and implementation

The first phase of process chain design is the accomplishment of the requirement analysis. Thereby, the product (form, shape, surface, etc.) is analyzed in order to define the requirements for production. The next phase of process chain configuration concerns the generation of process chains by using defined process modules. Several different process chains can be generated, tuned and afterwards evaluated. The incremental change in some parameters and process modules allows the optimization of the process chain scenario. For the optimization of the process chain, the technical feasibility, cost efficiency and the aspired quality have to be taken into account. Thereby, historical data, simulation techniques or experiments can be used as data sources. The best process chain scenario is chosen for implementation. After determining material flows and layout of the process chain resources the sequence will be constructed and commissioned. The phase of process calibration tries to improve the product quality of the process chain overall by adjusting single process modules with each other. At least, however, the production can be started.

10.3.3 Configuration and Evaluation of Process Chains

The configuration and evaluation of process chains is an important part of the design process. The configuration of process chains comprises many opportunities and possible scenarios, which have to be evaluated and compared to each other. Therefore it is necessary to have a systematic approach for the configuration of optimized process chains. *Denkena* describes a method for the configuration of process chain with three different configuration strategies [10]. The strategies of adoption, substitution and integration are depicted in Fig. 10.28.

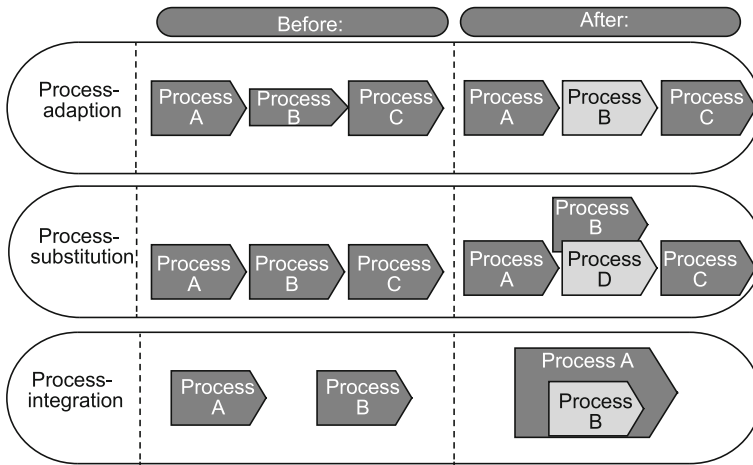


Fig. 10.28 Strategies for the configuration of process chains [11]

The first strategy of adaption tries to tune process steps in a way that they would better fit their predecessor and successor. In the figure it is shown that process B is adapted to processes A and C. That could be for example the adjustment of some process parameters in order to harden the workpiece for the next process step. The strategy of substitution is very useful, if there are other manufacturing technologies which are much more cost-efficient, technological feasible or produce a better product quality. An example could be the replacement of grinding (Process B) by hard turning (Process D). The strategy of integration is about the reduction of the operation sequence. Usually, the additional handling operations between the processes are eliminated by having more than one process in one machine. This can be done, for example, by a machining center. These strategies of configuration help generate process chain scenarios. In order to identify the most promising ones and to discontinue the worst ones, it is necessary to continuously evaluate the process chain scenarios, because it is planned from the outline to the detail. Ideally, each configuration step could be evaluated and compared to the other scenarios, but this approach is too time-consuming. Instead, the different scenarios are compared to each other by using the same level of detail.

The evaluation of process chains depends on three general requirements:

- process chains have to be technologically feasible
- process chains have to be cost-efficient
- process chains have to assure a high quality of manufacturing.

Obviously, the general requirements are in contrast with each other. Cost-efficient process chains almost never assure the best quality. Thereby, a tradeoff between technological demands, cost-efficiency and product quality have to be found. This can be done by the use of weighting factors and an objective function in order to determine the total value of across these three criteria:

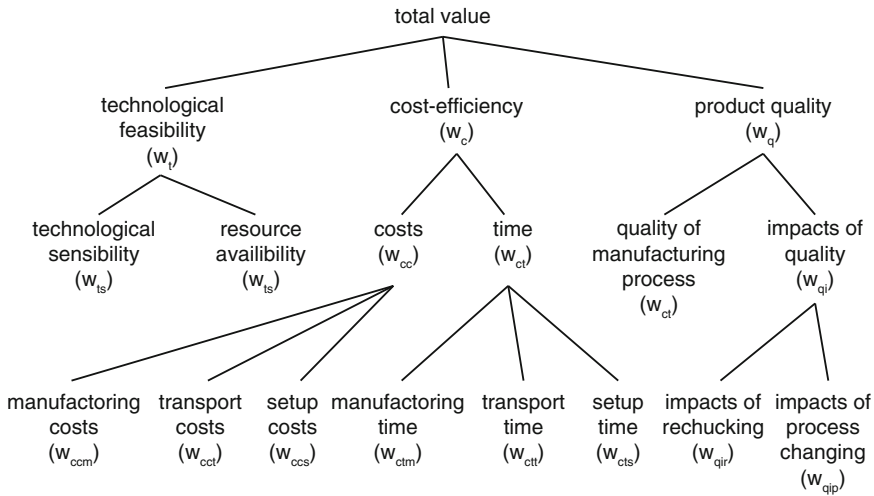


Fig. 10.29 Example of an objective tree for the evaluation of process chains [27]

$$totalvalue = W_t * P_t + W_c * P_c + W_q * P_q \quad (10.18)$$

W_t Technological weighting factor

P_t Technological evaluation parameter

W_c Cost efficiency weighting factor

P_c Cost efficiency evaluation parameter

W_q Quality weighting factor

P_q Quality evaluation parameter

One suitable method for evaluating different objectives in a structured way is the use of an objective tree. An objective tree starts with one over-arching objective, which is subdivided into several main objectives. These main objectives can also be divided in some sub-objectives and so on. This process is continued until a “tree” is created that covers all the separate objectives. In order to achieve a structured weighting of the different objectives, the sum of the weighting factors for each level is 100 %. For the first level it would be:

$$100 \% = W_t + W_c + W_q \quad (10.19)$$

W_t Technological weighting factor

W_c Cost efficiency weighting factor

W_q Quality weighting factor

To give an example, *Jurklies* used this method for the evaluation of process chains in the context of tool and mold making [26]. Tool and mold making demand high quality process chains with few deviations, which makes such a systematic

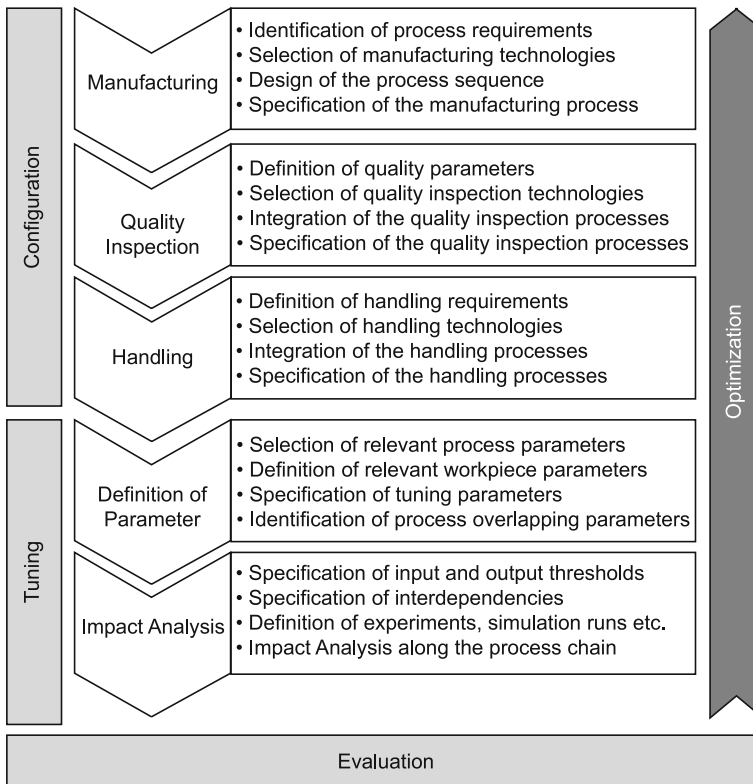


Fig. 10.30 Procedure for the generation and configuration of integrated micro manufacturing process chains [42]

approach very necessary. An enhanced structure of *Jurklies'* evaluation scheme is given in Fig. 10.29.

It is obvious that the evaluation of new process chains is much more difficult than the evaluation of process chains with minor changes. In particular, the estimation of technological feasibility and possible product quality requires a lot of experience and the support of knowledge-based planning systems.

10.3.4 Design Aspects of an Integrated Micro Manufacturing Process Chain

In comparison to the planning of macro process chains, the consideration of size effects implies an additional planning factor. Although the procedure of process chain planning outlined at the beginning of this chapter can also be used for micro process chains, the planning is much more challenging as the process chain behavior is more dynamic and complex than in the macro world. As a result, the

evaluation of the technical feasibility, cost-efficiency and product quality is much more difficult. Therefore, it is necessary to rely on a knowledge-based planning approach, in order to enable the quality-oriented design of micro process chains. The general idea is to make use of past experiences. The objective of the quality-oriented process chain is to identify serious issues early during the planning phase. Thereby, the focus is on the configuration and the evaluation of integrated micro process chains. The overall procedure for the configuration and tuning of integrated micro process chains is shown in Fig. 10.30. In order to enable an effective process design, this procedure should be followed in conjunction with the product design. Thereby, for each of the products features, suitable processes have to be selected and configured and tuned. As a result, dependencies between processes can be determined during the product design and serious issues can be taken into consideration early on. For instance, a specific manufacturing technology can be suitable to satisfy a specific production process, whereas its characteristics can lead to problems in another process later in the chain.

By considering manufacturing as well as quality inspection and handling processes, an integrated micro process chain will be designed. The first step is to identify process requirements and select adequate manufacturing technologies. Then the design of the process sequence starts by specifying the selected manufacturing processes. The next step integrates quality inspection processes into the process chain. Thereby, quality inspection processes will be defined based on specific quality parameters derived from global requirements, and of requirements implied by preceding manufacturing processes.

Afterwards, suitable quality inspection technologies will be identified or developed and will be integrated into the existing process chain of the manufacturing processes. In the same way, handling processes will be included in the process chain, depending on the handling requirements of manufacturing and quality inspection processes. After defining the integrated processes chain, the phase of tuning starts. Thereby, relevant input and output process parameters as well as the workpiece parameters will be defined and actual values will be assigned depending on the technologies defined. In addition, process tuning parameters are specified and process overlapping parameters are identified. Process overlapping parameters are either input or output parameters of two or more processes. As a result, these indicate dependencies between distinct processes within the overall process chain. These parameters could either refer to process specific characteristics (e.g. a manufacturing or quality inspection process requiring a minimum positioning precision from the preceding handling process) or they could refer to product specific characteristics (e.g. a manufacturing step requiring a specific roughness, which has to be ensured by the preceding manufacturing steps).

After defining the process chain's parameters, an impact analysis will be performed, in order to tune the process chain in an optimal way. By specifying input and output thresholds as well as rules for the interpretation of these thresholds, the interdependencies between the processes are defined. Additionally, experiments or simulation runs can be developed, in order to acquire data for sensitive processes.

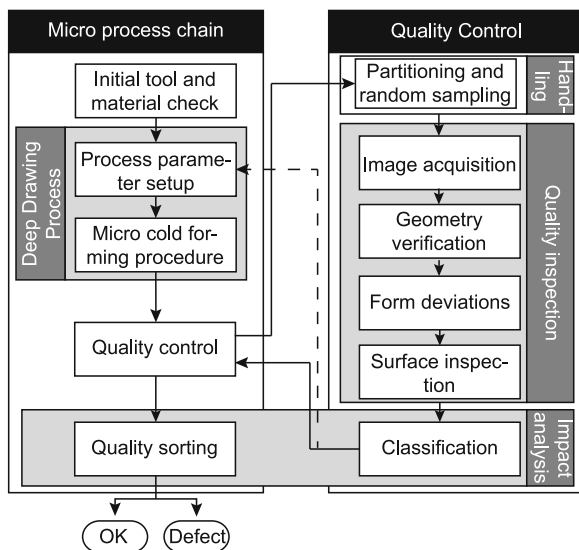
Finally, impact analysis will be performed along the complete process chain. It aims to identify issues between the different processes. The objective of the tuning phase is to solve all serious issues by adjusting the parameters, and therefore the selected technologies. The next step is to evaluate the process chain and compare the results with the other scenarios. The most promising scenarios will be detailed further and strived to be optimized. After some iteration, the best scenario will be identified and used for the general process design’s implementation phase.

10.3.5 Example: Quality-Oriented Design of a Micro Deep Drawing Process Chain

This section is about the specific quality-oriented design of a micro deep drawing process chain. Therefore an integrated process chain consisting of a quality inspection process and a micro deep drawing process is designed. This simple scenario shows some very relevant aspects of integrated quality control by presenting the methods applied for sampling and surface inspection. The structure of the process chain for micro deep drawing is shown in Fig. 10.31.

The left side of the image denotes the process chain, whereas the right side depicts quality control. The process chain starts with an initial tool and material check. During the deep drawing process, all relevant parameters for the manufacturing of a micro cup, e.g. drawing force and punch velocity, are set. At the beginning, all parameters are set to initial values relevant to the process window. After that, the operative cold forming procedure is performed. Following the manufacturing process, the procedure of quality control is executed. As micro cold

Fig. 10.31 Quality-oriented design of a micro deep drawing process chain



forming enables the bulk production of micro components, it is usually impossible to inspect all manufactured micro parts. This requires statistical methods to ensure reliable and accurate quality control. Instead of inspecting all micro components, quality inspection is applied to a subset of micro parts. Due to the deviation in processing speeds of the applied technologies, the first operation of the quality control process is the *partitioning and random sampling* of manufactured components. Finding an appropriate size and interval for these samples is a crucial step in statistical quality control. Large sizes and small intervals correspond to higher inspection costs. Each sample as well as the correlating subset requires additional handling and storage operations. Smaller sizes and longer intervals are cheaper, but could lead to substantial quality issues. Therefore, it is necessary to organize sampling in an efficient way.

10.3.5.1 Determining the Sample Size and Interval

In many manufacturing scenarios a constant sample size and interval is not sufficient, because there are long phases of very stable processes and only short phases of strong varying product quality. The idea of dynamic quality control is to adjust the sample size and interval depending on the actual process stability. Techniques like the sequential probability ratio test enable flexible, appropriate and reliable sampling for quality inspection. Another common technique for flexible subset quantity is fuzzy control. Fuzzy control systems are based on fuzzy logic, introduced by Zadeh [53]. Compared with classical Boolean logic systems, fuzzy logic provides the opportunity for a continuous transition between the belonging and not belonging of a statement to a specific set. Thereby, statements that are classified as true or false can only be processed to a certain degree. The objective is to dynamically adapt the size of the sample subset within the statistical quality control dependent on the quality index. Figure 10.32 illustrates the components and the operating principle of the fuzzy control system.

Within fuzzy control, the first step is the fuzzyfication of the input variables. In this case, the QI is represented as a linguistic variable. By using linguistic variables, information that is expressed linguistically can be modeled and processed without any loss of information. In the example, the range of the values of the QI is from 0 to 1. Within the fuzzy control, these values are expressed by the linguistic

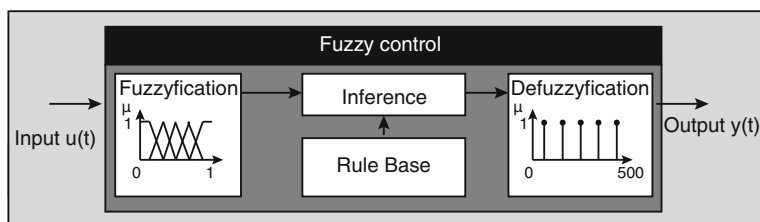


Fig. 10.32 Fuzzy control for determining the dynamic sample size

values *very bad*, *bad*, *normal*, *good* and *very good*. In order to transform the real value of the QI, the fuzzy control performs fuzzyfication. The fuzzyfication step in Fig. 10.32 shows the behavior of all membership functions. Each defined linguistic variable has its own membership function that describes to what extent the statement of the linguistic variable is true. After fuzzyfication the linguistic variables are analyzed by using a rule base that is generated by expert knowledge. For each single rule of the rule base the degree of fulfillment is computed. By analyzing all rules the inference step of the fuzzy control system is performed. The following defuzzyfication step generates the real value of the output value, in this case the size of the sample subset. There are many different ways of performing this step. In order to implement an efficient defuzzyfication, the fuzzy output variables are often modeled as so-called singletons. By using the gravity methods for singletons a real value for the size of the sample subset can be computed.

10.3.5.2 Surface Inspection and Rectification

All micro parts from the correct and significant subset are used for the *Quality inspection* sub-process. An image acquisition technique successfully applied in the micro domain, which satisfies the strict accuracy requirements of the image processing operation, is confocal laser microscopy [23]. This technique generates 2D and 3D information from the surface and form of the micro part. At the beginning of the *Quality inspection* sub-process, the focus is set to the verification of geometry and form deviations as described in the previous sections. In addition to these investigations, another important part of quality control is automated surface inspection. Here, image processing and statistical methods are used to find surface defects. A major challenge in automated surface inspection is the arbitrary shape, size and orientation of surface defects. Therefore, the surface inspection framework starts with a decomposition of the input image in different scales and orientations, using multi-scale analysis methods like the steerable pyramid technique [46]. After decomposition, the local binary pattern operator (LBP) is applied [37]. This operator is based on local pixel thresholding and calculates weights in a range

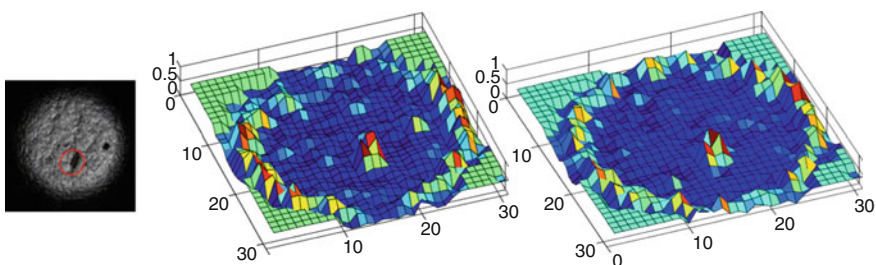


Fig. 10.33 Defect detection result of the surface inspection framework; from left to right: input image with labeled defect, two different statistical features

Table 10.7 Main defect classes during deep drawing

Defect	Example	Cause	Retification
Bottom wrinkling		Loose punch/Blank holder clearance	Reduce blankholder pressure
Wall wrinkling		Loose punch/Blank holder clearance	Reduce stamp radius
Bottom fraction		Overcharge drawing force	Reduce blankholder pressure, increase stamp radius

of [0, 255], where 0 corresponds to no change and 255 to a maximum change in texture. To detect the exact localization of a defect each LBP image is decomposed in overlapping blocks. In each block statistical values are calculated, like mean, median, and standard deviation. For comparison, all statistical values are normalized into a range of [0, 1]. A final decision on defect location results from a non-maxima suppression and merging of all individual block statistics. A detailed overview of the surface inspection process developed in CRC 747 is given in [43]. A result of the framework without segmentation of the micro cup border is shown in Fig. 10.33:

The figure shows an input image with a labeled defect. The two resulting images represent two different statistical features. The magnitude of the statistical results in the images correspond to the probability of a defect region. In this case, both results vote for the correct defect location.

Besides the detection of defects or form deviations, the classification of these imperfections is a major part of quality inspection. A detailed statistical analysis of different kinds of defect is important for further quality control and parameter adjustment. Defects occurring during deep drawing processes can have several reasons, like faulty manufacturing, material deficiency or thermal stress. A detailed overview of defects during deep drawing is given in [47]. Table 10.7 shows two main defect classes projected from the macro to micro domain. The micro cup has a diameter of 500 μm and was acquired using a confocal laser microscope. Further methods for optical measurement are described in Sect. 10.2.4.

The last step in the quality control process is *Impact analysis*. During this operation all defects occurring are evaluated. Significant accumulations of specific defect classes result in changes to the process parameters. The evaluation of the whole manufacturing process results in a new parameter setting, which is propagated back into the manufacturing process.

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