Investigation of Topography, Adhesion and Diffusion Wear in Sliding Contacts during Steel and Titanium Alloy Machining

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Abstract

The aim of the present thesis work is to increase the fundamental knowledge of the tribological contact between the cutting tool and the work material in three different cutting operations, i.e. hard milling of cold work tool steels, turning in 316L stainless steel and turning in Ti6Al4V alloy, respectively. The influence of cutting parameters and tool surface topography on the initial material transfer tendency and resulting wear and wear mechanisms were investigated under well controlled cutting conditions. High resolution scanning electron microscopy (SEM) and surface analysis, including energy dispersive X-ray spectroscopy (EDS), Auger electron spectroscopy (AES) and time-of-flight secondary ion mass spectrometry (ToF-SIMS), were used in order to characterize the worn cutting tools on a sub-µm scale and deepen the understanding of the wear mechanisms prevailing at the tool / work material interface. The characterization work includes the analysis of worn tool surfaces as well as cross-sections of these. Also, the back side of collected chips were analysed to further understand the contact mechanisms between the tool rake face and chip.

The results show that the transfer tendency of work material is strongly affected by the surface topography of the rake face and that an appropriate pre- and post-coating treatment can be used in order to reduce the transfer tendency and the mechanical interaction between a coated cutting tool and 316L stainless steel. The continuous wear mechanisms of the cutting tools were found to be dependent on the work materials and the cutting parameters used. In hard milling of cold work tool steels, polycrystalline cubic boron nitride shows a combination of tribochemical wear, adhesive wear and mild abrasive wear. In the turning of 316L stainless steel and Ti6Al4V alloy, using medium to high cutting speeds/feeds, the wear of cemented carbide is mainly controlled by diffusion wear of the WC phase. Interestingly, the diffusion wear processes differ between the two work materials. In contact with 316L stainless steel crater wear is controlled by atomic diffusion of W and C into the passing chip. In contact with Ti6Al4V crater wear is controlled by the diffusion of C into a transfer work material layer generating a W layer and TiC precipitates which repeatedly is removed by the passing chip. The experimental work and results obtained illustrates the importance of in-depth characterization of the worn surfaces in order to increase the understanding of the degradation and wear of tool materials and coatings in metal cutting operations.

Keywords: Tribology, Metal cutting, Cemented carbide, PCBN, CVD and PVD coatings, Surface topography, Wear mechanisms, Diffusion

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urn:nbn:se:uu:diva-390315 (http://urn.kb.se/resolve?urn=nbn:se:uu:diva-390315)
To Kamran and Arian

The two loves of my life
This thesis is based on the following papers, which are referred to in the text by their Roman numerals.


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Authors’ contribution to the publications

Paper I  Part of planning, major part of experimental work, part of evaluation and writing.

Paper II  Part of planning, part of experimental work, major part of evaluation and writing.

Paper III  Major part of planning, experimental work, evaluation and writing.

Paper IV  Major part of planning, experimental work, evaluation and writing.

Paper V  Major part of planning, experimental work, evaluation and writing.

Paper VI  Major part of planning, experimental work, evaluation and writing.
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1 Introduction

Cutting of metal is a topic of great interest for today’s industrial production. A cutting process refers to the process which removes the work material in form of chips by using for example turning, milling, boring or drilling, to achieve desired geometry of the work piece. The methods investigated in the present work are hard milling and turning.

The machinability of a material is how easily it can be machined using a cutting tool. Machinability may be assessed by one or more of the following criteria: tool life, limiting rate of metal removal, cutting forces, surface finish and chip shape [1]. Ideal conditions for cutting are long tool life, short cutting time and high cutting accuracy. In order to achieve these conditions, selection of efficient cutting conditions and tools, based on work material and machine capability is essential.

A high quality cutting tool must display the following characteristics:
- High hardness, especially at high temperatures
- High toughness (cracking and chipping resistance)
- High wear resistance

Cutting tool materials can be divided into the following material classes [1];
- High speed steels with the hardness of HV 800-1000 kgf/mm².
- Cemented carbides with the hardness of HV 1200-2000 kgf/mm².
- Ceramics with the hardness of HV 1500-2300 kgf/mm².
- Super hard materials (cubic boron nitride and diamond) with the hardness of HV 5000-10000 kgf/mm².

In the present work cutting tools of poly crystalline cubic boron nitride (PCBN) is used in milling cold work tool steel. PCBN is known for excellent mechanical properties such as high temperature strength, and hardness second to only diamond. Therefore, PCBN cutting tools are a good candidate for cutting hardened steel. Wear mechanisms of a PCBN cutting tool used in hard milling of cold work tool steels is investigated as a part of this thesis.

Although PCBN tools are capable of reaching higher cutting speeds in machining Ti-6Al-4V than cemented carbide tools [2], they are not popular in machining titanium due to their high price which can be 10-20 times higher than their cemented carbide counterparts [3].

Cemented carbides for metal cutting applications consist of more than 80% of the hard phase WC. The mechanical properties of cemented carbide are
controlled by amount of WC, binder phase and WC grain size. Hardness of cemented carbide increases by higher amount of WC phase and the finer grain size means higher hardness at a given binder phase content. Researchers have studied the machinability of titanium alloys in the past [4-6]. They suggested that straight cemented carbide (WC-Co) with Co content of ~ 6 wt.% and a WC grain size between 0.8 to 1.4 μm gives the optimum performance.

Titanium alloys are the most promising materials for the products of the spacecraft, aircraft and biomedical parts because of their high specific strength, fracture resistance, and corrosion resistance. Part of this thesis is dedicated to study machining of titanium with cemented carbide cutting tool of 6 wt.% Co content and WC grain size of 1 μm.

Austenitic stainless steels are used in a large number of different applications and industries, e.g. in the food processing, chemical, petrochemical, pulp and paper industries. Austenitic stainless steels are almost exclusively machined using CVD and PVD cemented carbide tools.

In order to increase the wear resistance of the cemented carbide tools for milling and turning thin coatings of TiAlN, Al₂O₃ and T(C, N), may be deposited on the surface by chemical vapour deposition (CVD) or physical vapour deposition (PVD) techniques.

Coatings produced are of a very high quality; they are homogenous, contain few defects (depending on substrate treatment) [7]. It has been shown that surface topography of a CVD and PVD coating may originate from three different contributions; roughness of the substrate surface, intrinsic coating characteristics such as grain size, phase composition or texture and surface defects such as macro particles or droplets. The role of surface topography of the coated cutting tools on the tribological behaviour (friction coefficient, transfer layer formation and wear) in machining austenitic stainless steels is studied as the other part of this thesis.

Although coated tools are used in machining austenitic stainless steels, a deep crater on the rake face and a large wear scar on the flank face will often involve a breach of this coating. Then knowledge about chemical dissolution or mechanical wear of the underlying substrate could be beneficial for both development of future substrate grades and for more detailed modeling of volumetric wear rates, considering both coatings and substrates.

Part of this thesis is dedicated to study wear mechanisms of uncoated cemented carbide cutting tools with different WC gain sizes in machining of austenitic stainless steels.
1.1 Research objectives

Cutting tool wear has a decisive role in any metal cutting process. It will affect the tool life, the robustness of the cutting operation and the quality of the machined surface. Consequently, understanding the contact condition prevailing at the tool/work material interface and the mechanisms controlling tool wear is of outmost importance. Tool wear is the result of the actual tribosystem and involves many complex chemical, physical and thermo-mechanical phenomena resulting in different wear mechanisms. The main goal of this thesis is to increase the understanding of the tribological degradation of cutting tools used in the machining of steel and titanium alloy. The work is based on well controlled cutting tests and post-test characterization of the worn tools and the produced chips using different types of microscopy and surface analysis techniques. The work is based on well controlled cutting tests, e.g. orthogonal turning tests using a flanged work piece, see Fig. 1, and post-test characterization of the worn tools and the produced chips using different types of microscopy and surface analysis techniques.

![Figure 1. Test set-up for orthogonal turning tests using a flanged Ti6Al4V work piece.](image)

The main objective of the thesis is to identify and explain a number of mechanisms occurring in sliding contacts during steel and titanium alloy machining, i.e.

- The dominant wear mechanisms of PCBN in the hard milling of cold work tool steels.
- The influence of tool surface topography on the material transfer tendency in the turning of 316 stainless steel.
• The dominant wear mechanisms of cemented carbide cutting in the turning of 316 stainless steel.
• The effect of WC grain size on wear of cemented carbide cutting tool in the turning of 316 stainless steel.
• The dominant wear mechanisms of cemented carbide in the turning of Ti6Al4V.
Metal cutting industry has been a crucial part of fabrications industry and the machine working. Metal cutting represents a dominating group of processing technologies such as turning, boring, shaping, milling, drilling, and broaching. In general, most research has been focused on the development of more high performance cutting tools, giving a better understanding of how an increase in quality and reduction in both processing time and costs can be obtained. Cutting processes can be classified as continuous or intermittent. In continuous cutting operations, cutter continuously remains in contact with the workpiece and the heating occurs in a steady linear fashion. Turning and boring are example of continuous processes. In intermittent process such as milling or shaping, a cutting edge is successively engaged and disengaged in a periodic way [8].

2.1 Turning

Turning is one of the most common machining processes. A part is rotated while a single point cutting tool is moved against the workpiece [9]. Turning produces cylindrical or conical surface and can be done manually in a traditional form of lathe or by using a computer numerical control (CNC) lathe machine.

In turning, three often considered cutting parameters are cutting velocity \( v_c \), feed rate \( f_n \) and depth of cut \( a_p \) whereas, tool life, cutting force and surface roughness is correlated with these parameters. Cutting velocity \( v_c \) is the rate at which the uncut surface of the work passes the cutting edge of the tool and in this thesis specified in terms of m/min. The feed rate \( f_n \) is measured as mm/revolution and it is the distance moved in the axial direction for longitudinal turning and in the radial direction for face turning. The depth of cut \( a_p \) is the width/depth (depending on cutting operation) of the metal removed from the bar and is specified in mm.

The cutting speed and the feed rate are the most important parameters which can be adjusted to achieve optimum cutting condition. The cutting tool which is used in turning needs to have high hardness even at the elevated high temperatures. A low friction surface is also desired to reduce the heat generation.
2.2 Hard milling

In milling, the cutting tool is rotated while the work piece is clamped on a table and the feed action is obtained by relative movement between the work piece and the cutter, often by moving the table. Typical cutters in milling operation have number of teeth vary from one to over one hundred, thus milling is a faster and higher productive process compare to turning. In milling process, each milling cutter tooth removes its share of the stock in the form of small individual chips. An important consequence of milling operations is that each edge is subjected to periodic impacts as it makes contact with the work piece material. Thus, it is stressed and heated during the cutting part of the cycle, followed by a period when it is unstressed and allowed to cool [1]. The milling of steels in the hardened state (45 HRC and higher), is termed hard milling.

2.3 The tool/work material contact

The formation of the chip by shearing action at the shear plane is the aspect of metal cutting which has a high importance in machining. The major part of the energy the cutting process generates is converted to heat through the produced friction and plastic deformation. Most of this energy is consumed in three deformation zones shown in Figure 2. The primary deformation zone (P₁) is the zone where the chip is formed. In the secondary shear zone (P₁I) the chip slide across the rake face of the cutting tool and this is the zone where the highest temperatures and pressures are obtained. In the tertiary deformation zone (P₁II), new surface of the workpiece is generated [8].

![Figure 2. The three deformation zones (P₁-P₁I) in orthogonal cutting configuration [8].](image-url)
2.3.1 Tool/chip interface

In cutting process, unwanted work piece material is removed to obtain a designed shape. The relative movement at the chip-tool rake face is conducted under high temperature, high pressure and high strain.

Movement of the chips across the rake face of the cutting tool cannot be treated as a classical friction situation in which the coefficient of friction is dependent only on the sliding force \( F_T \) and the force \( F_N \) normal to the sliding interface and independent of the contact area.

\[ \mu = \frac{F_T}{F_N} \]

is a useful measure only when the normal force on the apparent contact area is low compared with the yield stresses of the materials. When the normal force is increased to such an extent that the real area of contact is a large proportion of the apparent contact area, the real area of contact becomes independent of the normal force, and the frictional force becomes that required to shear the material across the whole interface. This force is almost independent of the normal force, but is directly proportional to apparent area of contact, a relationship directly opposed to that of classical friction concepts [1].

The phenomena occurring on the tool rake face depend on the local conditions of stress, velocity and temperature and on the local properties of tool and workpiece material. Under extreme conditions of pressure and temperature, chemical reactions between the components may occur which may change the known properties of the partner materials significantly.

2.3.2 The tool/work material interface

According to Astakhov [10] the contact between the tool flank and the workpiece machined surface occurs due to the spring back caused by the plastic deformation of this surface in cutting. As it is shown in Figure 3, the radial cutting force \( F_T \) in a face turning process causes the plastic deformation of the machined surface. After the cutting edge passes over, this surface recovers by \( \delta_1 \) causing the contact with the tool flank up to point A. This elastic recovery is proportional to the radial cutting force, which in turn, depends on the uncut chip thickness \( t_1 \).
2.4 Cutting tool materials

2.4.1 Cemented carbides
Cemented carbides are a range of composite materials, which consist of hard carbide particles sintered in a metallic binder. The carbides are often made of tungsten, titanium or niobium and the binder matrix consists of a metal mainly cobalt or nickel. Both coated and uncoated cemented carbide tools are used in machining. Typical application of the uncoated grades are machining of super alloys and titanium alloys. Coated cemented carbide tools have unique combinations of wear resistance and toughness and are extensively used in machining of steels. In the present study uncoated and coated cemented carbide are used as cutting tools in turning Ti6Al4V and 316L stainless steel, respectively.

2.4.2 Polycrystalline cubic boron nitride
Polycrystalline cubic boron nitride (PCBN) cutting tools have high temperature stability, high wear resistance and suitable mechanical properties for machining of hard and wear resistant ferrous materials. The tools are frequently used in machining of hardened bearing steels, case hardened steels and hardened cold and hot work tool steels. PCBN is a composite material comprising cubic boron nitride (cBN) grains in a binder matrix. These materials are produced by a powder metallurgical process and commonly categorized as high and low cBN content PCBN grades. High cBN content grades contain approximately 80-95% cBN in a metallic-type binder. Low cBN content grades contain from 40-70% cBN and the majority have ceramic based binder systems such as TiC and TiN [11].

Since high cBN content PCBN grades have higher hardness and fracture toughness than low cBN content PCBN grades, the former is a better choice for milling, but in finish hard turning, low content cBN grades usually gives
longer tool life and produces better surface finish [11, 12]. Wear of PCBN tools seem to be influenced by the tool geometry, cBN content, binder phase and hardness, composition and microstructure of the work material and cutting conditions [13, 14]. This complexity makes a great challenge when attempting to study the wear mechanism. Different wear mechanisms such as abrasion, adhesive wear, diffusion and tribochemical wear are involved in the wear of PCBN cutting tools.

2.5 Coatings

Vapour deposited coatings for metal cutting tools have been successfully used since the introduction of CVD TiC coated cemented carbide inserts 50 years ago. Since then various single and multilayer chemical vapour deposited (CVD) and physical vapour deposited (PVD) coatings of various wear resistant materials (mainly carbides, nitrides and oxides) have been developed. From the beginning, the focus of the coating development was on wear protection against mechanical wear. By increasing production rate, the focus of research shifts towards high temperature properties of coatings and their function as wear resistance layer. Nowadays, approximately 85% of all cemented carbide tools are coated [15].

Substrate material can be designed for strength and toughness while the coating is responsible for resistance to wear and thermal loads. Coatings act as a chemical and thermal barrier between the tool and workpiece [16]. Physical Vapour Deposition (PVD) and Chemical Vapour Deposition (CVD) are the two common types of thin-film coatings in the metal forming industry and have therefore been used in this study.

2.5.1 CVD coatings

At the present time, coatings obtained by the chemical vapour deposition are widely used in the cemented carbide industry. CVD technique produces well adhered, uniform, and dense surface layers which is high resistant against wear in cutting operations. CVD is a process conducted at elevated temperature (900-1100 °C). The process involves four stages; formulation of the reactant vapour, transport of this vapour, chemical reaction between the vapour and the heated substrate and finally removal of by-products [17]. In the CVD process volatile compounds are added to the chamber supplying the metal and non-metal constituents of the coatings. For example, to produce coatings of TiC, titanium tetrachloride (TiCl4) vapour is used to provide the Ti atoms, and methane (CH4) may be used to supply the carbon atoms for the coating [1]. During the process at high temperature, it is evitable to produce large re-
residual thermal stress inside the coating/substrate system. Different thermal expansions between the coating and the substrate can induce extrinsic residual stresses. This may result in many microcracks in the coating layer when the coating/substrate system is cooled down.

The combinations of TiC, TiCN, TiN, Al₂O₃, provide a considerable increase in tool lifetime of the coated cemented carbides. In this study, the tribological properties of CVD coated inserts have been investigated in papers II and III.

2.5.2 PVD coatings

PVD covers a broad family of vacuum coating processes which involve atomization of material from a solid source and deposition of that material to the substrate [18]. PVD coatings can be categorized according to the means of atomizing the source material and are often divided into evaporation and sputtering. Evaporation involves the thermal vaporization of the source material and sputtering is a process in which the source material is made cathodic and is bombarded with ions. Cathodic arc evaporation, electron beam evaporation and cathodic sputtering are the most common PVD techniques. PVD was developed almost 10 years later than CVD [7]. The thickness of PVD coatings is controllable, so sharp coated edge can be achieved. PVD processes are carried out at lower temperatures, compare to CVD process. The lower deposition temperature of PVD, results in lower residual thermal stress in the coatings compared with CVD. PVD coatings are usually characterized by high residual stresses in their structure due to the thermoelastic effects generated by particles of high kinetic energy during deposition, then compressive stress which inhibits the crack growth are another beneficial properties of PVD coatings [19]. In the present work tools used in paper II, and III have been coated using a cathodic arc evaporation PVD process.

2.5.3 Comparison between PVD and CVD

Main characteristics of PVD and CVD deposition techniques are illustrated in Table 1.
Table 1. Comparison of coating process characteristics used in the coating of cemented carbide cutting tool.

<table>
<thead>
<tr>
<th>Process characterization</th>
<th>PVD</th>
<th>CVD</th>
</tr>
</thead>
<tbody>
<tr>
<td>Method</td>
<td>Process in a vacuum chamber $10^{-4}$ - $10^{-9}$ Torr [20]</td>
<td>Processed in vacuum or at atmosphere pressure chamber</td>
</tr>
<tr>
<td>Temperature</td>
<td>Low process temperature 200-500 ºC</td>
<td>High process temperature 900-1100 ºC</td>
</tr>
<tr>
<td>Deposition</td>
<td>Physical</td>
<td>Chemical</td>
</tr>
<tr>
<td>Typical thickness range</td>
<td>1-5 μm</td>
<td>4-12 μm</td>
</tr>
<tr>
<td>Advantages</td>
<td>+Tough edge for intermittent operations such as milling. +Sharp edges for fine cutting and finishing. +High built in compressive stresses, allow resisting crack initiation.</td>
<td>+Good adhesion +Possibility of thick coating +High resistance against wear and oxidation +Appropriate for a higher cutting speed</td>
</tr>
<tr>
<td>Disadvantages</td>
<td>- Less good adhesion</td>
<td>-Brittle eta phase (low toughness) - Limited thickness - Cooling cracks due to built-in tensile stress.</td>
</tr>
</tbody>
</table>

2.5.4 Pre- and post-coating treatment

In order to improve the performance of a coated cutting tool, a pre- and post-coating treatments of the cutting edge and rake face region are frequently used. Surface treatment used, cover microblasting, brushing and polishing to improve the surface roughness of the cutting tool. The surface smoothing processes reduce the friction and the tendency to work material pick-up. Besides, the use of microblasting of CVD coatings, frequently showing a tensile stress state, has the potential to reduce the tensile stress state and tendency to cracking.
3 Tribology in metal cutting

3.1 General aspects
Tribology is the science and technology of interacting surfaces in relative motion, which includes the entire field of friction, wear and lubrication. Friction and wear are complex phenomena which are controlled by a variety of different mechanical, physical, chemical and thermal processes active at the interface between the mating surfaces. The tribological conditions in metal cutting are extreme and controlled by the mechanical and thermal stresses active at the tool/work material contact zone and very high stresses, strains and temperatures can be obtained. A considerable amount of heat is transferred into the cutting tool and work piece but in the case of machining work materials having low thermal conductivity the heat concentrates on the cutting edge.

Therefore, contact length between the tool and the chip affects cutting conditions and performance of the tool and tool life [21]. A contact of tool-chip interface results in crater forming in the rake face of cutting tool, while the contact of tool clearance face-work material results in a flank wear land flattening of the clearance face. Figure 4 illustrates the wear by comparing initial condition and steady state condition at rake and clearance faces.

![Figure 4](image)

Figure 4. Schematic drawings of contact condition of the rake face/chip and clearance face/work material interface at a) initial condition and b) steady state condition.
The wear of a cutting tool is rather complex and involve different wear mechanisms. Also, when cutting tools are coated with multi-layered coatings the complexity increases. During the initial cutting tool wear stage, the coating will control the wear rate. For a new cutting edge, the coating play important role. One important factor which can improve the performance of the coating in this stage is surface topography [22-25]. During this stage, tool life (coating life) can be increased by reducing the topography of the coating by a post-coating treatment, e.g. grit blasting, brushing, polishing. Also, micro grit blasting can be used to reduce the tensile stress in CVD coatings to reduce the tendency to cutting edge failure due to chipping. A smoother coating surface will reduce the mechanical interaction between the coating and work material. Chemical composition and microstructure (e.g. WC grain size) of cutting tool substrate are important factors influencing the wear rate of the substrate [26, 27]. Figure 5 shows the cross section of a worn coated cemented carbide insert. As can be seen, the coated insert displays a large crater formed in the rake face and removal of the coating in the flank area.

Tribology research in understanding metal cutting is mainly focusing on the design and development of new coatings and tool materials in order to increase quality and longer tool life.

![Figure 5. Cross section image of a worn coated cemented carbide cutting tool illustrates the presence of flank and crater wear (the flank and crater wear formed in clearance and rake face respectively). Note that the crater is filled with transferred work material.](image)

### 3.2 Wear and wear mechanisms

In general, tool failure can be the result of three different degradation processes;
i) A gradual continues wear due to the tribological interaction between the tool and the work material.

ii) Macroscopic plastic deformation of the cutting edge (change of the geometrical shape).

iii) Macroscopic mechanical brittle fracture of the cutting edge.

In order to achieve a robust metal cutting operation, degradation processes ii) and iii) must be avoided by selecting an appropriate cutting tool and appropriate cutting parameters (cutting speed, cutting feed, depth of cut) optimized for the selected work material. Consequently, a gradual wear of the cutting tool is preferred since it leads to a robust machining operation and a long tool life. Gradual wear generally occurs at two different locations on a cutting tool, i.e. the rake face and the flank wear land. In general the gradual wear is said to be controlled by the following wear mechanisms;

- Abrasive wear
- Adhesive wear
- Diffusion/dissolution wear (tribochemical reactions)

Usually in turning, abrasive wear is more common at the flank face where the temperature is lower. Tribochemical wear, which is a thermally activated process, is a common wear mechanism of the crater wear, where the maximum cutting temperature occurs. Adhesive wear can occur along the entire contact zone and be caused by both thermally activated and micro-mechanical process.

3.2.1 Dominant tool wear and failure mechanisms

Most common material removing mechanisms controlling the wear of cutting tools in turning are adhesive, abrasive and tribochemical wear. The dominant wear mechanism of the cutting tool depends on the cutting conditions and the material of the both tool and workpiece.

*Abrasive wear* occurs by abrasive action of hard particles in the work material [28]. These particles can be the different hard phases of the work material or some strain-hardened fragments of the built up edge. Also, tool wear fragments may being pulled-out from the cutting edge resulting in self abrasion of the cutting tool.

*Adhesive wear* is caused by the mechanical removal of the tool material when the adhesive junctions are broken [29]. Workpiece material adheres on the cutting tool surfaces; shearing force results in fracture of junctions and some small fragments of the cutting tool or coating, breakaway which results in deterioration of the tool surface, but at very high temperature, tribochemical wear takes place.
Tribochemical wear means the wear is of a chemical nature, where reaction occurs in the interface of cutting tool/work material/chip. In machining, layers of work material might adhere on the rake and flank face of the cutting tool. Chemical species can diffuse from the tool surface towards the adhered layer and vice versa. Tribochemical reactions are activated by mechanical interactions between the contacting surfaces whereas activated surface sites and localized high temperature is sufficient for chemical reactions. Hence, the tribochemical reactions such as diffusion or dissolution are always in combination and interaction with other wear mechanisms and is not a wear mechanism alone. The chemical diffusion changes the contact conditions and facilitates the tool failure.

3.2.2 Macroscopic wear types

The typical macroscopic wear criterion for cutting tools is crater wear, flank wear, notch wear, plastic deformation or chipping. The first 3 wear types are illustrated in Figure 6.

![Figure 6. Schematic of a worn cutting tool showing crater, flank and notch wear](image)

**Crater wear:** Crater wear consists of a crater or a groove on the tool rake face formed by the action of the chip sliding on the surface, which reduces the load bearing capacity of the tool. Crater wear is often measured by a profilometer as the maximum depth of the crater or the volume loss of material from the rake face of the tool.

**Flank wear:** Flank wear occurs on the tool flank face as a result of friction between the machined surface of the workpiece and the tool. Cutting forces increase significantly with flank wear and if the amount of flank wear exceeds some critical value, the excessive cutting force may cause tool failure [30].
Notch wear: Notch wear occurs by the rubbing of the machined surface with the cutting tool at the depth of cut line. The machine surface may develop a thin work hardened layer, this contact could contribute to notch wear [31].

Plastic deformation: Plastic deformation takes place when the tool material is softened due to the high cutting temperature and is exposed to high cutting forces at the same time. Since elevated temperature significantly decreases the yield stress of cutting tools, at high cutting temperature, plastic deformation is inevitable.

Chipping: Brittle failure occurs due to high contact stresses at the cutting edge as a result of combination of critical cutting parameters. The weak cutting edge due to the crater also contributes to the brittle failure.

3.3 Machined surface

A machining process produces a surface characterized by the topography, metallurgy and mechanical properties of the machined surface. These surface aspects indicate that a machined surface consists of interrelated features that influence the surface functional performance [32].

Within a machining process, chemical, mechanical, and thermal mechanisms will always be present. A high-energy input increases the likelihood of metallurgical damage, i.e. micro-crack, microstructure deformation and plastic deformation. Mechanical effects of the machining processes change the surface finish and microstructure and causes plastic deformation. Thermal effects causes phase change and a change in the dislocation density. The combination of these, affects the surface integrity of the machined surface, i.e. friction and other abrasive effects of the machining generates local heating which causes plastic deformation and phase changes, so residual stresses and consequently cracks may form on the machined surface. Chemically affected layers are generated by the surface chemical changes resulting from a combination of the mechanical and thermal influences.

In turning, surface finish and dimension tolerance are affected by:

i) factors due to machining parameters, such as feed rate, cutting speed, and depth of cut,

ii) factors due to cutting tool parameters, such as tool wear, tool geometry, tool material and tool coating,

iii) factors due to machining and machine tool conditions, such as dry or wet turning, type of cutting fluid, method of fluid application, machine tool rigidity, chatter vibration and

iv) factors due to workpiece material properties, such as hardness, microstructure, grain size and inclusions [33].
4 Surface characterization techniques

In order to fulfil the main research objective of the thesis work a number of microscopy and surface analysis techniques have been used in the characterisation of the tribo surfaces. The used techniques includes optical interface profilometry, Scanning Electron Microscopy (SEM), Energy Dispersive X-ray Spectroscopy (EDS), Auger Electron Spectroscopy (AES) and Time of Flight Secondary Ion Mass Spectrometry (ToF-SIMS). In the following sections the techniques will be presented.

4.1 Optical interference profilometry

Measurements with optical interference profilometry provides detailed information regarding the surface of a sample i.e. 2D profile and 3D pictures as well as topographical parameters describing the surface. This profilometry technique utilizes a dual-LED illumination source to measure nanometre surface roughness to millimetre craters. Two working modes are available: VSI (Vertical Shift Interference) and PSI (Phase Shift Interference). The VSI filter uses a combination of green and white LED illumination. The VSI technique which uses a broadband light source, is dedicated to roughness measurement up to 1 mm topography and PSI technique which uses a narrowband green light source, is typically used to test smooth surfaces (roughness less than 30 nm) [34]. Different surface parameters such as $R_a$, $R_z$ and $S_k$ are provided with this device. The $R_a$ value is the average height of deviation of peaks and valleys from the mean line over the sampling length, see Figure 7.

$$R_a = \frac{1}{l} \int_0^l |y(x)| \, dx$$
Figure 7. The schematic of roughness measurement showing $Ra$ and $Rz$ calculation.

$Rz$ is calculated by averaging the measured vertical distance from the highest peak to the lowest valley within five sampling lengths.

\[
Rz = \frac{R1 + R2 + R3 + R4 + R5}{5} \quad - \quad \frac{R6 + R7 + R8 + R9 + R10}{5}
\]

The Skewness ($SSK$) parameter is used to describe the difference in symmetry of the surface profile around the mean line. For example, an as-deposited PVD coating with high amount of droplets would result in $SSK > 0$ while the value for a polished one with shallow craters would result in $SSK < 0$.

In present thesis, Wyko NT9100 in VSI mode was used in order to measure the crater wear volume formed on the rake face of the cutting tool (Papers I, IV, V and VI) and to investigate the pre-coating substrate surface topography as well as the topography of the coated samples (Papers II-III).

4.2 Scanning Electron Microscopy (SEM)

The scanning electron microscope uses a focused beam of high-energy electrons to generate different signals at the surface of a specimen. Electrons are emitted from an electron source such as a W or a LaB$_6$ filament or a Field Emission Gun (FEG). Electrons are accelerated in the electron gun in the range of 0.1-30 keV. The electrons are focused with aid of magnetic lenses to a fine electron beam. With the aid of magnetic coils, the electron beam is scanned in a rectangular pattern over the sample surface. When the electron beam hits the surface, the electrons interact with the atoms in the sample surface and can be elastically or inelasticity scattered, so the primary electrons travel down to a limited depth in the sample resulting in an activated volume.

The most important signals which provide useful information are secondary electrons (SE), backscattered electrons (BSE), Auger electrons and characteristic X-ray photons. Mostly either the SE or BSE signals is used to produce
SEM images. Secondary electrons are electrons of the specimen ejected during inelastic scattering of the energetic beam electrons [35]. Secondary electrons have low kinetic energy and are generated within a thin layer beneath the surface, therefore imaging with SEs gives high resolution, showing morphology and topography of the sample surface. BSEs are produced by an elastic scattering interaction of the beam electron and atom nuclei. Backscattered electrons are recognized by high kinetic energy. Since the heavier elements produce more BSEs in the same image, region with higher atomic number appears brighter, therefore BSEs are used for illustrating atomic number contrasts in multiphase samples. High acceleration voltage provides the high contrast, imaging with BSEs. In the case of lower accelerating voltage, a better quality of the surface and fine-scale structures will be produced with SEs. In Figure 8 the principles of SEM is illustrated. In this thesis, SEM (FEG-SEM, Zeiss Ultra 55) has been used to characterize worn cutting tools, coatings topography and material transfer tendency of different surface treated coatings.

![Figure 8. The schematic showing the principles of SEM [36].](image)

### 4.3 Energy Dispersive X-ray Spectroscopy (EDS)

When the beam electron interact with tightly bond inner shell electrons of a specimen atom, an electron from the shell will eject, so the atom is left in an excited state. The atom relaxes to its ground state through an allowed transition of outer shell electrons filling the inner shell vacancy. The excess energy
will then be emitted as an X-ray photon which is the characteristic for each element [35] or can be given to a second, emitted electron, an Auger electron.

In order to detect the emitted characteristic X-ray photon, the most SEM devices are equipped with an Energy Dispersive Spectrometer, EDS. By using EDS the elemental composition of the sample will be determined. The information depth depends on the activated volume and activated volume is dependent to the voltage and chemical composition of the sample, i.e. the activated volume increases with increasing accelerating voltage and increasing atomic number decreases the activated volume. Figure 9 shows the activated volume and detected electrons and radiation. In this thesis, EDS technique (Oxford Inca Energy) has been used to analyse the element composition of the worn surfaces.

![Figure 9. Activated volume and the emitted signals [37].](image)

### 4.4 Auger Electron Spectroscopy (AES)

Auger Electron Spectroscopy (AES) is one of the most surface sensitive analytical techniques for determining the composition of the 10-40 Å surface layer of sample specimen. When the atoms in the specimen are bombarded with the primary electron beam, at the first step by removal of a core electron, atomic ionization occurred and an ion in an excited state is produced. At the second step an L-electron jumps and fill the vacancy and at the last a KLL Auger electron is emitted in order to conserve the energy released in step 2 [37, 38], see Figure 10. Analysis of the emitted Auger electrons makes it possible to obtain compositional information of the surface.
There are four different modes of operation in AES: Area analysis (point analysis and larger areas), line scan analysis, mapping and depth profiling. In area analysis, a total spectrum is acquired from the selected area on the sample surface, often in the energy range of 20-2000 eV. Auger line scan analysis is acquisition of data along a line in the surface. The electron beam is stepped point-by-point along the line and each point is analyzed in an energy interval around peaks of interest. The number of points decide the analysis time.

In mapping the electron beam is scanned over a selected area of the sample. The Auger intensity is measured at each point of the area. The image displayed shows the peak intensity of each pixel and the resolution (number of pixels) decides the analysis time. The most common is 128x128 pixels for each image.

Auger depth profiling provides depth distribution of elements or thickness of a surface layer using a combination of argon ion sputtering. One cycle of a typical depth profile consists of sputtering to remove a layer of material, stopping, measuring relevant portions of the Auger spectrum, and performing elemental quantification.

In the present work (Papers I, IV, V and VI) diffusion layer forms at the tool-chip interface is characterized using AES (Ulvac-PHI 700 Xi Scanning Auger Nanoprobe).

4.5 Time-of-Flight Secondary Ion Mass Spectrometry (ToF-SIMS)

Time-of-Flight Secondary Ion Mass Spectrometry (ToF-SIMS) is one of the most powerful surface analysis methods in terms of high sensitivity, high spatial resolution imaging, and detailed chemical information. For a ToF-SIMS analysis, a solid surface is bombarded by primary ions of some keV energy to remove particles from the surface of the sample. ToF-SIMS is a very surface
sensitive technique because the emitted particles originate from the uppermost one or two monolayers. Particles could be positive or negative ions produced closer to the site of impact or molecular compounds farther from the impact site. The particles are accelerated into a "flight tube" and their mass is determined by measuring the exact time at which they reach the detector (i.e. time-of-flight). ToF-SIMS mass spectra allow the identification of all constituent elements, isotopes, and molecular species by their mass-to-charge ratios [39]. ToF-SIMS is a widely used technique for depth profiling applications in many areas. For depth profiling, Dynamic ToF-SIMS conditions must be applied. Dynamic SIMS describes the removal of one or more atomic/molecular layers per analytical cycle. Each data point in a depth profile thus represents an analytical cycle. In this thesis (Paper IV and VI) the analyses in the PHI TRIFT II instrument were performed with a pulsed liquid metal ion gun (LMIG) with a source enriched in $^{69}$Ga isotopes.
5 Experimental work and main results

The experimental part of the thesis work was based on well-controlled machining tests and post-test characterization of the worn tools and produced chips. In the following section, details regarding the machining tests and the work and tool materials used will be given. The machining tests covered the machining of three different work materials, i.e. hard milling of cold work tool steels, turning of 316L stainless steel and turning of Ti6Al4V alloy.

5.1 Work piece materials

Cold work tool steels are commonly used for a wide variety of cold work applications, e.g. sheet metal forming and powder pressing. In order to meet the requests for tools used in these applications, i.e. a high hardness, sufficient toughness and high wear resistance, cold work tool steels are highly alloyed and contain a significant amount of hard carbides which contribute to the high hardness/high wear resistance. Unfortunately, the high amount of hard carbides makes cold work tool steels difficult to machine.

In paper I the hard milling of two different powder metallurgy (PM) cold work tool steels, i.e. Vancron 40 and Vanadis 4 Extra, see Table 2, were used for the milling experiments. The microstructure of Vancron 40 consist of a matrix of tempered martensite with V(C, N) carbonitrides with hardness of 2800 HV and (Mo, W)C carbides with hardness of 1500 HV. Vanadis 4E consists of a matrix of tempered martensite, and (V, Mo, Cr) C carbides with hardness of 2800 HV [40].

Table 2. Chemical composition (wt.%), hard phase content and hardness of the cold work tool steels investigated.

<table>
<thead>
<tr>
<th>Steel grade</th>
<th>C</th>
<th>N</th>
<th>Si</th>
<th>V</th>
<th>Cr</th>
<th>Mn</th>
<th>W</th>
<th>Fe</th>
<th>Hardness (HV3)</th>
<th>Hard phase content vol.%</th>
</tr>
</thead>
<tbody>
<tr>
<td>Vanadis 4E</td>
<td>2.9</td>
<td>0.5</td>
<td>9.8</td>
<td>8</td>
<td>0.5</td>
<td>1.5</td>
<td>-</td>
<td>Bal.</td>
<td>810±30</td>
<td>MC 8 %</td>
</tr>
<tr>
<td>Vancron 40</td>
<td>1.1</td>
<td>1.8</td>
<td>0.5</td>
<td>8.5</td>
<td>4.5</td>
<td>0.4</td>
<td>3.2</td>
<td>Bal.</td>
<td>790±30</td>
<td>MC, 5% MCN, 19 %</td>
</tr>
</tbody>
</table>

Austenitic stainless steels are the most widely used stainless steel grade known for their formability and resistance to corrosion. However, when it comes to machining, austenitic stainless steels are regarded as materials that are difficult to machine [41, 42]. This is mainly due to their low thermal conductivity,
high work hardening tendency, strong bonding to the cutting tool and poor chip breaking properties, all of which will contribute to a pronounced tool wear and reduced tool life.

The work material used in papers III-IV was a molybdenum-alloyed austenitic stainless steel with improved machinability, Sanmac® 316L, see Table 3.

### Table 3: Chemical composition (wt.%) and hardness of the Sanmac® 316L austenitic stainless steel work material.

<table>
<thead>
<tr>
<th>C</th>
<th>Si</th>
<th>P</th>
<th>S</th>
<th>Cr</th>
<th>Mn</th>
<th>Ni</th>
<th>Mo</th>
<th>V</th>
<th>Hardness, HV₀.₁</th>
</tr>
</thead>
<tbody>
<tr>
<td>≤0.03</td>
<td>≤0.75</td>
<td>≤0.04</td>
<td>≤0.03</td>
<td>16.5</td>
<td>≤2.0</td>
<td>11</td>
<td>2.1</td>
<td>≤0.05</td>
<td>235</td>
</tr>
</tbody>
</table>

**Titanium and its alloys** are generally classified as “difficult to machine materials” due to typical combinations of its chemical, mechanical and thermal properties and its chip formation characteristics [43-46]. This is mainly due to a combination of properties including:

- the high temperature strength of titanium and pronounced work hardening ability has a negative impact on the chip formation mechanisms
- the low heat conductivity of titanium promotes high temperatures at the cutting edge and the rake face
- the segmented nature of the chips generates cyclic stresses which promotes surface fatigue of the cutting edge
- the small chip/tool contact length and the very thin flow zone between the chip and rake face result in high stresses and high temperatures in close connection to the cutting edge
- the high chemical reactivity between titanium and the tool material promotes strong bonding and diffusion wear at the rake face resulting in a crater.

The work material used in the paper V and VI was a mill-annealed bar of Ti6Al4V alloy, grade 5, supplied by Timet, see Table 4.

### Table 4. Chemical composition (wt.%) and hardness of the Ti6Al4V alloy according to the material specification provided by material supplier Timet.

<table>
<thead>
<tr>
<th>Al</th>
<th>V</th>
<th>Fe</th>
<th>C</th>
<th>O</th>
<th>N</th>
<th>Y</th>
<th>Ti</th>
<th>Hardness, HV₅</th>
</tr>
</thead>
<tbody>
<tr>
<td>6.425</td>
<td>3.970</td>
<td>0.155</td>
<td>0.019</td>
<td>0.190</td>
<td>0.006</td>
<td>&lt;0.001</td>
<td>Bal.</td>
<td>300</td>
</tr>
</tbody>
</table>
5.2 Cutting tests

5.2.1 Hard milling of Cold work tool steels

Hard face milling (climb milling) experiments in paper I were performed using round PCBN inserts (ISO-Code RDHW10T3 M0S-01030-LF, chamfer length: 0.10 mm, chamfer angle: -30º, edge radius: 15 mm) and a commercial milling equipment (Mori Seiki SV500). The experiments were performed under dry cutting conditions using a 50 mm diameter cutter body equipped with one single PCBN insert. The PCBN tool used in the study was a high cBN-content PCBN grade with 85 vol.% cBN (grain size 2 µm) in a binder phase of W, Co and Al (15 vol.%). Cutting parameters for the main tests series were cutting speed $v_c$: 100 and 150 m/min, feed $f_z$: 0.1 mm/tooth, cutting depth $a_p$: 0.5 mm, cutting width $a_e$: 12.5 mm. In order to evaluate the flank and crater wear overtime, milling experiments were run for 10, 20, 30 and 40 min in the two cold work tool steels.

5.2.2 Turning of stainless steel

For the turning tests with coated cemented carbide cutting tools, a radial turning process involving the tool tip was used. The machining experiments were performed using commercial TCMW 16T304 insert grades GC 2025 and GC 1115 by Sandvik Coromant at wet condition. The cutting velocity, $v_c$, was 180 m/min and cutting times, $t = 1$ and 10 minutes.

For the turning tests with uncoated flat-faced cemented carbide inserts (ISO code: TCMW 16T304), the turning was performed in an orthogonal direction using a flange cutting process with a pre-prepared workpiece with flanges 3 mm wide and 3 mm apart, and with a 25 mm depth along the cylindrical surface. Turning tests were performed under wet conditions at cutting velocity $v_c = 150, 200$ and 250 m/min and cutting time, $t = 12$ seconds.

The turning tests, were performed using a commercial George Fischer CNC lathe, NDM-17/125. Depth of cut, $a_p$, was 2 mm, the radial feed rate, $f_n$, was 0.2 mm /rev, and the cutting fluid used in these experiments was a solution of Hocut 4160.

5.2.3 Turning of Ti6Al4V

Orthogonal turning tests were performed using uncoated flat-faced cemented carbide inserts (ISO code: TCMW 16T304) and a commercial George Fischer CNC lathe, NDM-17/125. Cutting parameters for the main tests series were; cutting velocity $v_c$: 90 and 115 m/min, feed rate $f_n$: 0.05, 0.10, 0.15 and 0.2 mm/rev, cutting depth $a_p$: 3.0 mm. Besides, complementary tests were per-
formed also at lower cutting speeds, i.e. 30, 45 and 60 m/min. Turning experiments were run for 0.5, 1, 2 and 3 min. All tests were performed with a solution of a semi synthetic coolant, Hocut B50S.

5.3 Post-tests characterization

After the turning tests, the wear of the rake face, i.e. the crater wear volume, was measured by 3D optical surface profilometry, using a WYKO NT9100 optical surface profilometer. The flank wear, i.e. the width of the flank wear land (VB), was measured by scanning electron microscopy (SEM) and used to calculate the flank wear volume.

The worn cemented carbide surfaces were analyzed using a high resolution FEG-SEM (Zeiss Ultra 55) equipped with an Energy Dispersive X-ray Spectroscopy system (EDS; Oxford Instruments Inca). The FEG-SEM was also used to analyse the back side of chips, i.e. to analyse whether traces of cemented carbide, either as fine particles, a diffusion layer, or combinations of these, could be detected.

Auger Electron Spectroscopy (AES; Ulvac PHI 700 Xi Scanning Auger Nanoprobe) and Time of Flight - Secondary Ion Mass Spectrometry (ToF-SIMS; PHI TRIFT II), two very surface sensitive analyzing methods, were used to analyze the top-surface of the worn cemented carbide and the back-side of the produced chips in order to evaluate the presence of any tribo chemical layers or diffusion zones.

AES analysis was performed using an accelerating voltage of 10 kV and a primary electron beam current of 10 nA. Depth profiling was performed using 1.0 kV Ar+ ion sputtering with a sputter rate of 17 nm/min measured on a Ta2O5 reference sample with known thickness (100 nm). Computer software from PHI-Matlab was used to evaluate the Auger depth profile data. Linear least squares (LLS) fitting is used to extract different chemical states of the elements found.

ToF-SIMS analysis was performed with a pulsed liquid metal ion gun (LMIG) with a source enriched in 69Ga isotopes. Depth profiles were obtained by sputtering a surface area of 100 µm × 100 µm with a continuous non-pulsed beam with primary ion energy of 15 keV and an aperture giving a current of ~600 pA was used. Positive static SIMS mode was used to analyse an area of 50 µm × 50 µm in the centre of the sputtered area. All SIMS spectra were calibrated using peaks with known mass/charge ratio.
5.4 Wear of PCBN in hard milling of tool steels

Figure 11a shows an overview of a worn PCBN cutting tool edge after 40 minutes milling in the Vanadis 4E at $v_c$ 100 m/min. As can be seen cutting tool shows both pronounced crater and flank wear. At a higher magnification, see Figure 11b, the PCBN material displays a slightly scratched appearance with smooth scratches in the chip sliding direction (crater region) and work material sliding direction (flank region). Similar wear characteristics were observed for milling of the Vancron 40 grade.

![Figure 11a](image1)

![Figure 11b](image2)

*Figure 11. Wear characteristics of PCBN cutting tools after 40 minutes milling Vanadis 4E at a cutting speed of 100 m/min. Note the crater and flank wear regions.*

At a higher magnification, $>5000 \times$, the worn PCBN material, despite the multiphase structure, displays a smooth surface both in the flank wear region and the crater wear region, see Figures 12 and 13.

However, in both wear regions, a thin adherent film covers the worn PCBN material. In Figure 12, a representative region of the worn flank is imaged using two acceleration voltages. With the lower acceleration voltage (3 kV, left column) it is clear that the worn surface is to a large extent covered by a
thin adhered film. The adhered film appears irregular and is not connected to any visible surface damage or irregularity in the PCBN surface. With the higher voltage (10 kV, right column), the adhered film is noticed as an almost entirely transparent film revealing the multiphase microstructure of the PCBN material which displays a smooth surface with no evidence of grain pull-out.

![Figure 12](image1.png)

**Figure 12.** Characteristics of worn flank surface as observed in the SEM using two different acceleration voltages; 3 kV (a) and 10 kV (b). Note that exactly the same area is imaged in both micrographs in order to illustrate the effect of accelerating voltage on the interpretation of the formation of an adhered film. The sliding direction of the work material is from top to bottom. Milling time 40 min, $v_c$: 100 m/min, work material Vanadis 4E.

![Figure 13](image2.png)

**Figure 13.** Characteristics of worn crater surface as observed in the SEM. The chip sliding direction is marked with an arrow. Milling time 40 min, $v_c$: 100 m/min, work material Vanadis 4E.

Figure 14 shows AES depth profiles obtained from the worn crater region of the PCBN tool used for milling Vanadis 4E at 150 m/min. The AES depth profiles were all obtained from small areas, $0.5 \times 0.5$ mm$^2$, making it possible to study the interface between the adhered film and the individual phases of
the PCBN material, i.e. the cBN and the binder phase. As can be seen all areas analysed is covered by a thin Si₉Oₓ rich tribofilm. Also, the adhered film/binder phase interfacial zone is significantly broader as compared with the adhered film/cBN phase interfacial zone, especially in the case of the Co-rich binder phase, indicating a chemical interaction.

Figure 14. AES depth profiles showing the elemental composition of the adhered film formed in three areas as facing the cBN phase (a) W-rich binder phase (b) and the Co-rich binder phase (c) in the crater region after milling the Vanadis 4E grade. Cutting data; \( v_c \): 150 m/min and milling time 20 min.

SEM of the counter surfaces, see Figure 15 and 16 indicates that the hard phase particles are plastically deformed and sheared in the cutting direction and thus not act as abrasive particles. Instead, the scratches observed are believed to be the result of local plastic deformation caused by PCBN wear fragments, mainly cBN grains, being pulled-out from the cutting edge. Also, the fact that the flank and crater wear regions both display relatively smooth worn surfaces with very little signs of any preferential wear of the PCBN material phase constituents, i.e. the cBN phase and the metallic binder phase, indicates the presence of tribochemical wear in combination with adhesive wear and abrasive wear on a very fine scale.
Figure 15. Cross section of machined surface of Vanadis 4E (a) and Vancron 40 (b) (milling time 10 min, \( v_c \): 100 m/min). Note the plastic deformation and shearing of the M₆C carbides (white) and M(C, N) carbonitrides (grey) in the machined Vancron 40 surface.

Figure 16. Surface characteristics of chips (from Vancron 40 steel grade) surface using two different acceleration voltages: 3 kV (a) and 10 kV (b) in contact with the rake face of the PCBN tool showing a smooth oxidized appearance.

5.5 Material transfer in sliding contacts

It’s a well-known fact that coating topography strongly influences the tribological performance of the tribosystem under sliding contact conditions. This is further illuminated by the results in Papers II and III. In Paper II, the influence of different surface micro topographies, obtained by various pre- and post-coating treatments, on the initial material transfer of austenitic stainless steel in sliding contact with two CVD and PVD coatings has been investigated. The work has been performed using CVD coated with an interfacial bonding layer of Ti(C,N), on top of which there is a multilayer of Al₂O₃. The outermost layer is TiN which improves wear resistance and is used for wear detection.
The PVD coating used for the cutting tools in the present study is TiAlN-(Al, Cr)$_2$O$_3$.

The sliding wear tests were performed using a commercial scratch test equipment, CSM Revetest, where a steel pin of austenitic stainless steel is sliding against the coated sample using a normal load of 20 N, a sliding distance of 5 mm and a sliding speed of 10 mm/min, see Figure 17. CVD and PVD coated cemented carbide sample with various pre- and post-coating treatments were used as the samples.

![Figure 17. Photo of the scratch test set up.](image)

Figures 18 and 19 show the surface morphology of the as deposited CVD and PVD coatings deposited on polished substrates. For the CVD coating, the surface irregularities mainly originate from the coating growth process and the resulting crystal structure with a large number of sharp facets. Larger particle, up to 20 µm, are also found on the surface. The average number of these was found to be 28±5 / mm$^2$. For the PVD coating, the surface irregularities mainly originate from a relatively large number of droplets (particles) in the size range 0.2 – 1.0 µm and a lower number of larger particles, size > 3 µm.
Figure 18. Surface characteristics of the CVD coating investigated (deposited on a polished cemented carbide substrate). a) Intrinsic surface morphology. b) Typical macro-particle.

Figure 19. Surface characteristics of the PVD coating investigated (deposited on a polished cemented carbide substrate). a) Intrinsic surface morphology. b) Typical macro-particle.

Figures 20 and 21 show the friction and acoustic emission characteristics of the as-deposited and post-polished CVD and PVD coatings in sliding contact with the austenitic stainless steel. As can be seen, all as-deposited coatings, also those deposited on polished substrates, show a high friction coefficient. Obviously, the as-deposited coating micro topography and morphology have a very strong impact on the friction behaviour. In contrast, coating post-polishing has a significant positive effect when it comes to reducing the friction coefficient of the sliding contact. Consequently, a low surface roughness is of utmost importance in order to obtain a low friction coefficient. Also, assuming that the acoustic emission signal reflects the degree of tribological interaction between the mating surfaces (transfer of stainless steel to the coating) the acoustic emission signals illuminate a significantly lower interaction in the case of post-polished coatings. Finally, it should be noted that for the post-polished coatings, also the pre-coating substrate treatment has a significant impact on the friction behaviour, cf Figures 20 and 21. Consequently, in order
to minimize the friction both a pre-treatment of the substrate surface and a post-treatment of the deposited coating are of importance.

*Figure 20.* Friction and acoustic emission characteristics of as-deposited CVD (above) and PVD (below) coatings deposited on ground (left column) and polished (right column) cemented carbide substrates, respectively. The acquisition rate is 50 Hz.

*Figure 21.* Friction and acoustic emission characteristics of post-polished CVD (above) and PVD (below) coatings deposited on ground (left column) and polished (right column) cemented carbide substrates, respectively. The acquisition rate is 50 Hz.
Figure 22 shows material transfer layers on the different coatings as-deposited on ground (GAD) and polished (PAD) substrates. As can be seen, large parts of the sliding wear track is covered by transfer lumps of counter material.

![Counterbody sliding direction →](image)

Figure 22. Details of wear track with counter body material build up on CVD (above) and PVD (below) coatings deposited on ground (left column) and polished (right column) substrates, respectively. Counter body sliding direction from left to right.

Figure 23 shows material transfer layers on the post-polished coatings deposited on polished substrates (PP) and ground substrate (GP). As can be seen, the amount of transferred material is significantly lower in the case of polished substrates as compared with ground substrates.

The post-polished coatings display two different scenarios. Post-polished coatings deposited on ground substrates show limited improvement in tribological performance due to the presence of scratches in the surface originating from the ground substrate. Obviously, the limited polishing (corresponding to a removed layer of approximately 0.2–0.3 µm in thickness) of the coatings deposited on rough substrates is not enough to obtain a smooth surface which inhibits material transfer. Consequently, since the CVD and PVD coatings deposited on cutting inserts in many applications have a coating thickness of only 4–5 µm, grinding of the substrates prior to the coating deposition process must be considered in an industrial process. In contrast, post-polished coatings deposited on polished substrates show significantly
improved tribological performance (low friction, low material transfer). However, there is a significant difference between the CVD and PVD coatings related to their microstructure. The dense microstructure, free from visible defects, of the CVD coating promotes the possibility to obtain a smooth post-polished surface.

Counterbody sliding direction →

*Figure 23.* Details of wear track with counter body material build up on CVD (above) and PVD (below) coatings deposited on ground (left column) and polished (right column), respectively. Counter surface sliding direction from left to right.

The results of the scratch tests show that micro-topography of the coated surfaces is partly controlled by the substrate surface topography and partly by the intrinsic coating micro-topography. All these topographical features affect the sliding contact, i.e. the friction characteristics and the transfer of stainless steel to the harder coated surface. On CVD coatings the large particles are found but the surface roughness mainly originates from the crystal growth structure. On PVD coatings, the number of droplets makes it problematic to define the surface roughness. The defects are present throughout the entire depth of the coating, so even a polished PVD coated surface deposited on a polished substrate is not perfectly smooth and defect free which could inhibit material transfer. However, a smooth surface of CVD coatings is obtainable by post treating the coating, which is deposited on a polished substrate.
5.6 Correlation between laboratory tests and cutting tests

In paper III, the impact of tribological testing in metal cutting in association with tribological response of mating surfaces and the influence of tool surface topography on the initiation and build-up of transfer layers is investigated. The results show that the transfer tendency of work material is strongly affected by the surface topography of the rake face. For both types of inserts, the initial transfer and the build-up of transfer layers are localised to microscopic surface irregularities on the rake face. Consequently, an appropriate surface treatment of the cemented carbide substrate before coating deposition and polishing CVD and PVD coating can be used in order to reduce the transfer tendency and the mechanical interaction between the mating surfaces. The coating-substrate systems showing a rough initial surface (most pronounced for the as-deposited coatings on fine ground substrates) also displayed a significant amount of discrete micro chipping of coating fragments resulting in a rougher surface and the presence of adhered work material, see Figures 24 and 25.

Figure 24. Details of the surface morphology in rake face showing the transfer of work material and surface failure of a) as-deposited CVD coating on fine ground substrate and b) post-polished CVD coating on polished substrate (bright areas correspond to adhered work material). Note the high tendency of adhered work material (bright areas) and coating micro chipping in a).
Figure 25. Details of the surface morphology in rake face showing the transfer of work material and surface failure of a) as-deposited PVD coating on fine ground substrate and b) post-polished PVD coating on polished substrate. Note the high tendency of adhered work material (bright areas) and coating micro chipping in a).

EDS analysis of the worn surface reveals, besides the presence of adhered work material, the presence of a thin transfer film rich in O, Si, Al and Ca, see Figure 26. The transfer film was more evident on the smoother coating surfaces, i.e. on post-polished coatings on polished substrates, and especially on the CVD Al₂O₃-layer. In contrast, adhered transferred work material was found to dominate on the rougher coating surfaces.

Figure 26. EDS spectra (E₀ =10 keV) obtained from the worn surfaces in Figures 24b and 25b corresponding to; a) transferred work material (bright spots) and b) formed transfer film on CVD Al₂O₃ layer (dark areas), and c) transferred work material (bright spots) and d) formed transfer film on PVD (Ti,Al)N layer (dark areas).

An improved surface finish was found to reduce coating wear and consequently the crater wear rate of the inserts investigated. Table 5 shows the wear volumes of craters formed on CVD- and PVD-coated inserts with different pre- and post-treatments after 10 min turning in the 316L stainless steel. This can most likely be explained by the reduced tendency to discrete chipping of
coating fragments in the contact zone and the formation of a thin transfer film composed of Al, Si, Ca, O with beneficial friction properties which are promoted by a smooth coating surface.

It should be noted that a direct quantitative comparison between the CVD- and PVD-coated inserts can not be done based on the present results, due to the differences in coating chemical composition, coating thickness and cemented carbide substrate grade. For example, the higher wear rates for the PVD-coated inserts could be caused by faster wear through the thinner (Ti,Al)N-(Al,Cr)2O3 coating and/or a faster wear of the more fine grained cemented carbide substrate. The influence of cemented carbide microstructure (WC grain size) on crater wear in the turning of 316L stainless steel will be treated in section 5.7. However, the present work established that both types of coatings show improved (lower wear rate) after the performed surface treatments.

Table 5. Crater volumes of the inserts investigated after 10 min turning tests. The scatter of values are ±0.1 10⁶ µm³.

<table>
<thead>
<tr>
<th>Coating</th>
<th>As-deposited on fine ground substrate</th>
<th>Post-polished on fine polished substrate</th>
</tr>
</thead>
<tbody>
<tr>
<td>CVD</td>
<td>2.2 10⁶</td>
<td>1.2 10⁶</td>
</tr>
<tr>
<td>PVD</td>
<td>11.8 10⁶</td>
<td>5.9 10⁶</td>
</tr>
</tbody>
</table>

5.7 Wear and wear mechanism of cemented carbide in the turning of 316L stainless steel

Three different uncoated cemented carbide grades, 462, H13A and 570 (Sandvik Coromant designation), with different mean WC grain sizes (fine, medium and coarse) were used in the turning of 316L stainless steel.

The chemical composition and hardness of the cemented carbide grades are given in Table 6. The chromium carbide in the fine-grained material is added by the manufacturer to inhibit grain growth during sintering, and the γ-phase (Ta,Nb)C in the coarse-grained material is added to improve hardness and chemical stability.

All inserts were flat without any geometry.
Table 6. *Cemented carbide grades investigated in the present study.*

<table>
<thead>
<tr>
<th>Grade</th>
<th>Chemical composition [vol.%]</th>
<th>Mean WC grain size [µm]</th>
<th>Hardness, HV3</th>
</tr>
</thead>
<tbody>
<tr>
<td>Fine-grained</td>
<td>WC 91.1, Cr$_3$C$_2$ 0.8, Co 8.1</td>
<td>0.3</td>
<td>1825</td>
</tr>
<tr>
<td>Medium-grained</td>
<td>WC 89.8, Co 13.2</td>
<td>0.8</td>
<td>1580</td>
</tr>
<tr>
<td>Coarse-grained</td>
<td>WC 89.6, (Ta,Nb)C 1.6, Co 8.8</td>
<td>1.0</td>
<td>1350</td>
</tr>
</tbody>
</table>

General turning of steels, including stainless steels, using cemented carbide inserts and recommended cutting parameters will result in crater and flank wear. Figure 27 shows an overview of a worn cemented carbide insert (fine-grained grade) after turning at $v_c$ 250 m/min for 12 sec.

![Image of worn cemented carbide insert](image)

*Figure 27.* General characteristics of worn fine-grained cemented carbide cutting insert showing crater and flank wear. Cutting time 12 sec, $v_c$: 250 m/min.

In common for all three cemented carbide grades is a relatively smooth worn surface, both in the crater and on the flank wear land, without any signs of micro chipping of wear fragments.

Figures 28 a and b show the influence of mean WC grain size on the resulting crater wear volume and flank wear volume after 12 seconds turning respectively. As can be seen while the crater wear displays an increasing wear rate with decreasing mean WC grain size, the flank wear displays a small decrease with decreasing mean WC grain size.
Figure 28. Influence of mean WC grain size and cutting speed on resulting crater wear volume (a) and flank wear volume (b) after 12 seconds turning of 316 L stainless steel.

Figures 29 and 30 show SEM images illustrating the worn surface morphology within the crater formed on the rake face of the three different cemented carbide inserts at two different cutting speeds, 150 m/min and 250 m/min, respectively. At the lower cutting speed (150 m/min) the worn surfaces are relatively smooth, not showing any pronounced topographical features. However, at the higher cutting speed (250 m/min) the worn surfaces show numerous ridges parallel with the chip sliding direction. As can be seen, the tendency to ridge formation increases with increasing mean WC grain size.

Figure 29. Characteristics of worn cemented carbide (after removing adhered work material by etching) within the crater after 20 sec turning at \(v_c\): 150 m/min. a) fine-grained b) medium-grained and c) coarse-grained cemented carbide. Chip sliding direction from left to right.

Figure 30. Characteristics of worn cemented carbide (after removing adhered work material by etching) within the worn crater after 12 sec turning at \(v_c\): 250 m/min. a) fine-grained b) medium-grained and c) coarse-grained cemented carbide. Chip sliding direction from left to right.
SEM and EDS analysis of the coarse-grained cemented carbide grade, containing gamma phase, revealed that the presence of (Ta,Nb)C grains in the microstructure were responsible for the pronounced ridge formation. The higher wear resistance of the (Ta,Nb)C grains make them protrude from the worn surface protecting the cemented carbide microstructure behind them (with respect to the chip sliding direction), see Figure 31. For the fine- and medium-grained cemented carbide grade, no microstructural features could be found explaining the formation of ridges in the surface.

Figure 31 a) Pronounced ridge formation in the crater of the coarse-grained cemented carbide insert after 12 sec turning at \( v_c \): 250 m/min. b) EDS spectrum showing the elemental composition of the first grains in the ridges. Chip sliding direction from left to right.

Figure 32 shows AES depth profiles obtained from the worn crater region of the coarse-grained cemented carbide grade after 12 seconds turning at a cutting speed of 250 m/min and illustrates the difference in surface elemental composition between smooth WC grains and WC grains showing a fine scale topography. While the smooth WC grain, see AES depth profile in Figure 32b, shows the presence of pure WC, the rough areas shows the presence of a thin layer with significant amounts of O, Cr, and Fe, which decreases with increasing sputtering depth, see Figure 32c. The gradual decrease of O, Cr, and Fe and the gradual increase of C and W with increasing depth indicate the presence of a tribochemical (diffusion) wear mechanism being active within the crater at high cutting speeds. It should be noted that both AES depth profiles were obtained from small areas, 0.3 × 0.3 \( \mu \text{m}^2 \), making it possible to study the interface between the adhered film and the individual WC grains in detail.

Complementary ToF-SIMS analysis from the worn crater region, showing a high number of WC grains covered by a thin tribo-layer supports the obtained AES results, i.e. the presence of Cr, Fe and W in the top-most surface, see Figure 33. These results are consistent with results obtained by Fieandt et. al. [47] who studied the chemical interactions between cemented carbide and 316L stainless steel using a diffusion couple method. It is believed that the suggested diffusion mechanism should be rate-limited by a reaction occurring at the interface between the WC phase and work material. Further insights
could be gained by identifying activation energies with the restriction of surface exchange rates or other possible interfacial reactions, applying the results to wear models similar to those in the works by Zanger and Schulze [48] or Malakizadi et al [49].

Figure 32. a) Detail of worn cemented carbide in the crater showing WC grains with different surface morphology. AES depth profiles showing the elemental surface composition of WC showing a smooth surface morphology (b) and WC grain showing a thin tribo-layer with a very fine-scale (nm-level) surface morphology (c). Cutting data: 12 sec turning at $v_c$: 250 m/min.
The flank wear lands of cemented carbide cutting tools used for turning of stainless steel show signs of plastic deformation of individual WC grains. This is revealed by visible slip lines in Figure 34. Besides, signs of superficial plastic flow can be seen on some of the WC grains.

The results obtained show that the WC grain size of the cemented carbides has a significant impact on the wear of the cemented carbide inserts in turning 316L austenitic stainless steel. The higher crater wear rate of fine-grained ce-
mented carbide at high cutting speeds is believed to be due to a higher interfacial micro contact area attributed to the smaller WC grain size which promotes a higher diffusion wear rate.

The slightly lower flank wear rate of the fine-grained cemented carbide grade at high cutting speeds is believed to be due to a higher mechanical strength (higher hardness) as compared with the coarse-grained cemented carbide which will reduce the tendency to wear caused by superficial plastic flow [27]. However, it can not be completely ruled out that presence of small amounts of Cr- or (Ta,Nb)-carbides affects the wear to some degree, but the observed trends are consistent through all three grades, suggesting that any such effects are small.

Figure 35a shows the ToF-SIMS depth profile performed on back side of the chip. The depth profile reveals a significant concentration of W in the topmost surface of the chip, which supports the above “diffusion wear” degradation mechanism of cemented carbide. Also, it shows that the oxidized chip surface is build-up of a three-layer oxide structure with an outer Fe-rich oxide, an intermediate W-rich oxide and an inner Cr-rich oxide, see Figure 35b. It should be noted that the lower sensitivity of AES (as compared with ToF-SIMS) made it difficult to analyse the diffusion layer using the AES technique.

![TOF-SIMS depth profile](image)

**Figure 35.** a) TOF-SIMS depth profile on backside of stainless steel chip (surface in contact with the rake face of fine-grained cemented carbide grade) illustrating the diffusion of W into the chip during the turning operation. b) Schematic showing the layered oxide structure of oxide film on back-side of chip. Cutting time 12 sec, $v_c$: 250 m/min.

Comparing chips produced by the fine- and coarse-grained cemented carbide inserts, see Figure 36, reveals that these observed fine scratches most probably are generated when the chip slides against the slightly worn cemented carbide surface within the crater wear zone, i.e. the size of the scratches is controlled by the size of the ridges in the worn surface. This
indicates that sliding mainly takes place along the chip/cemented carbide interface and not between the chip and an adhered layer.

Figure 36. Surface morphology of back-side of chips sliding against a fine-grained (a) and coarse-grained (b) cemented carbide insert. Cutting time 12 sec, \( v_c \): 250 m/min.

5.8 Wear and wear mechanisms of cemented carbide in the turning of Ti6Al4V

For the turning of Ti6Al4V, the cemented carbide grade H13A, a WC-Co grade which combines good wear resistance and toughness, was used. Figure 37 shows a worn cemented carbide cutting insert after 0.5 min turning in the Ti6Al4V alloy using a cutting speed of \( v_c \): 90 m/min, and feed rate of \( f_n \): 0.2 mm/rev. As can be seen, the cutting insert shows both crater wear and flank wear as well as adhered work material on the rake face and the flank wear land.
At low cutting feeds/cutting speeds, the worn cemented carbide within the crater displays a relatively rough worn surface underneath the adhered work material indicating that fragments of WC grains or, more probably, individual WC grains had been torn away in contact with the sliding chip, see Figure 38.

At higher cutting feeds/cutting speeds, corresponding to higher cutting temperatures at the rake face/chip interface, the worn cemented carbide within the crater displays a very smooth worn surface underneath the adhered work material, see Figure 39. As can be seen, the interface is very sharp without any porosity, demonstrating a strong adhesion between the work material and the cemented carbide, and reveals the formation of a thin, 100-150 nm, fine grained interfacial layer.
As it is discussed in detail in paper V, the most beneficial way to investigate diffusion layer is to perform an AES depth profile on an etched surface.

Figure 39 shows AES depth profiles obtained from the center of a crater formed on the rake face of a cemented carbide insert used for turning at high cutting speed after wet-etching. The fine scale surface morphology in Figure 40b mainly corresponds to a pure W-phase with a thickness of 50-60 nm after which the W-signal decreases and the C-signal increases reaching the expected composition of WC at a depth of approximately 130 nm. However, the AES depth profile also reveals the presence of TiC in the surface which suggests that C originating from the WC phase in the cemented carbide has diffused outwards and reacted with Ti forming TiC, see Figure 40c.

By using the linear least squares (LLS) routine two different components of the carbon depth profile can be distinguished, i.e. C in WC (C(W)) and C in TiC (C(Ti)), see Figure 41a. Figure 41b shows the C KLL Auger spectra.
from reference samples of TiC and WC, respectively. Red spectra corresponds to C in WC and blue spectra corresponds to C in TiC. The main Auger transition C KL\(_{23}L_{23}\) is positioned around similar kinetic energy, i.e. 276 eV for WC and 277 eV for TiC. The main difference between the C KLL Auger spectra can be seen in the fine structure between 250 and 270 eV where the C KL\(_{1}L_{1}\) and C KL\(_{1}L_{23}\) transitions show different energy positions. For TiC these Auger transitions are at 258 eV and 266 eV and for WC at 256 eV and 264 eV, which is in accordance with published data [50].

**Figure 40.** a) SEM image of an area in the center of the worn crater after wet etching (cutting speed: 90 m/min). b) AES depth profile obtained from area 1. Similar depth profiles were obtained from areas 2 and 3). c) AES depth profile obtained from area 4.

**Figure 41.** Two different C KLL Auger spectra extracted from a carbon depth profile. C in WC (red spectra) and C in TiC (blue spectra) b) Carbon signals from reference samples of WC and TiC.
Scratch testing of the WC$_{1-x}$ layer in the crater area revealed no tendency to adhesive failure. At low normal loads using a sharp diamond stylus (radius 7.5 µm), see Figure 42, post-test SEM studies of the scratches revealed no tendency to adhesive failure of the WC$_{1-x}$ layer. At low normal load, 0.01 N, the layer displayed a low cohesive strength where the individual columnar grains WC$_{1-x}$ layer tend to detach and smear out in the scratch track, see Figure 42c. A higher normal load, 0.05 N, see Figure 42d, results in a more macroscopic deformation of the cemented carbide but no tendency to extensive failure of the cemented carbide composite.

![Figure 42](image)

*Figure 42.* SEM images from the crater area of an etched insert illustrating the scratching characteristics of the WC$_{1-x}$ layer. a) Low magnification overview image showing three scratches within the crater. b) and c) Scratching response at a normal load of 0.01 N, d) Scratching response at a load of 0.05 N.

High resolution SEM images as well as high resolution AES depth profiling of the diffusion layer reveal the formation of an approximately 100-150 nm thick carbon depleted WC$_{1-x}$ layer and TiC precipitation at the cemented carbide/Ti6Al4V interface. Based on the present findings, it is believed that the diffusion of carbon from the WC-phase, resulting in a carbon depleted WC$_{1-x}$ layer with reduced mechanical strength, will promote detachment of the adhered build-up layers of work material on the rake and flank surfaces when
exposed to high shear stresses. If we assume that the WC$_{1-x}$ layer will stick to the detached build-up layers, i.e. that cracking will take place along the WC/WC$_{1-x}$ interface or within the WC$_{1-x}$ layer as soon as it reaches a critical thickness, the detachment of the build-up layer will expose the WC phase for the work material which immediately will adhere to the cemented carbide surface promoting the diffusion-controlled wear of the cemented carbide to continue. Consequently, the diffusion rate of carbon into the build-up layer and the removal rate of the WC$_{1-x}$ layer determine the wear rate of the cemented carbide tool material. Evidence of carbon depletion of the WC resulting in the formation of W layer and (Ti,V)C at the cemented carbide/Ti6Al4V interface in turning is reported in recent published papers [51-53]. The above wear model is supported by the results from detailed SEM, EDS and Auger analysis of the back-side of the chips which revealed the presence of Ti6Al4V build up layer fragments, 10-20 μm in diameter, see Figure 43, in the chip surface. At higher magnification, see Figures 43b-d, a high amount of fine debris originating from the diffusion layer can be seen embedded in the surface of the fragments. In contrast, the surrounding chip surface do not contains any wear fragments originating from the cemented carbide insert.

Figure 43 a) Fragments (indicated by arrows) of Ti6Al4V build-up layers containing fine debris of the diffusion WC$_{1-x}$ layer found on the back-side of Ti6Al4V chips. b) and c) High magnification SEM micrographs showing the presence of WC$_{1-x}$ debris in the surface of the detached build-up layer fragments. d) SEM tilted-view micrograph of the build-up layer fragment in Figure 43c.
Plastic deformation of the cutting tool is the result of a combination of high cutting temperature and high compressive stress. In turning titanium alloy, low thermal conductivity and high temperature strength of titanium leads to plastically deformation along the cutting edge. Figure 44 shows the plastic deformed cutting edge after 3 minutes turning Ti6Al4V with cutting parameters \( v_c \) 90 m/min, \( f_n \) 0.2 mm/rev.

![Plastic deformation of cutting edge](image)

*Figure 44. Cross section of worn coarse-grained cemented carbide cutting insert showing plastic deformation.*

High magnification SEM image in the plastically deformed region revealed that the as-sintered continuous WC skeleton is partly broken up and infiltrated by binder phase, see Figure 45. As can be seen, the Co binder phase has infiltrated the broken grain boundaries forming thin lamella and a more open microstructure.

![SEM images illustrating the degradation of the cemented carbide structure in the plastically deformed region 20 µm below the surface in the center of the flank wear land](image)

*Figure 45. SEM images illustrating the degradation of the cemented carbide structure in the plastically deformed region 20 µm below the surface in the center of the flank wear land). a) Undeformed WC-Co microstructure b) deformed WC-Co microstructure. Cutting speed: 90 m/min, feed rate: 0.2 mm/rev, time: 1 min.*

The results of the present study indicate that crater wear is controlled by mainly two different wear mechanisms. At combinations of lower cutting feeds and cutting speeds attrition wear will dominate while at combinations of
higher cutting feeds and cutting speeds diffusion wear will dominate. At high cutting feeds and cutting speeds plastic deformation, resulting in a depression of the cutting edge, and cracking and macroscopic chipping within the crater is frequently observed after 1-2 min. The wear and failure mechanisms of cemented carbide tools in the turning of Ti6Al4V is summarized in a wear map and it is presented in Figure 46.

**Figure 46.** Wear map of cemented carbide cutting tool after 2 minutes turning in Ti6Al4V using different combinations of cutting speed and feed rate.

### 5.9 Diffusion wear in the turning of 316L stainless steel and Ti6Al4V - A comparison

Figure 47a and b show cross sections of worn cemented carbide insert after turning 316L stainless steel and Ti6Al4V respectively. As can be seen, the worn cemented carbide, within the crater wear region displays a very smooth and well defined wear surface. After machining, a layer of work material adhered to the crater area. In the case of stainless steel no diffusion zone is observed at the interface of cemented carbide/adhered work material, see Figure 47a, while high magnification SEM image shows a thin diffusion zone at the cemented carbide/adhered work material interface in machining Ti6Al4V, see Figure 47b. However, in both cases a very smooth worn surface underneath the adhered work material indicates the presence of a continuous wear mechanism being active on a very fine, nm, scale.

It should be noted that comparison between machining stainless steel 316L and Ti6Al4V is done after same cutting length with different cutting parameters in order to reach cutting temperatures that have been estimated as reasonably similar.
The results of collected data from worn crater areas and backside of the chips in turning stainless steel 316L and Ti6Al4V, indicates that, in case of 316L stainless steel, diffusion wear of cemented carbide occurs through atomic diffusion into the chip, while diffusion wear of cemented carbide during turning Ti6Al4V occurs through atomic diffusion into the transfer layer, resulting in a diffusion zone (layer). Figure 48 shows a schematic of stainless steel 316L and Ti6Al4V chip sliding against cemented carbide surface.

Figure 48. Schematic of chip sliding against cemented carbide surface in turning a) stainless steel 316L and b) Ti6Al4V
Main Conclusions of thesis work

- Wear of the high cBN content PCBN cutting tool during hard milling of powder metallurgy cold work tool steels display crater wear, flank wear and edge micro-chipping. A combination of tribochemical reactions, adhesive wear and mild abrasive wear is believed to control the flank and crater wear of the PCBN cutting tool.

- The intrinsic micro-topography of as-deposited CVD and PVD coatings promotes material transfer. Material transfer is located to surface irregularities on the surface originating from the coating growth process. For the as-deposited CVD coating the nm scale topography of the crystals controls the transfer while for the PVD coatings the μm-scale droplets and craters control the transfer.

- Post-polishing of the coatings significantly improves the tribological performance of the surface reducing the friction coefficient and the material transfer tendency. For post-polished coatings, also pre-polishing of the cemented carbide substrate improves the tribological performance of the surface. The presence of μm-sized droplets and craters in the PVD coatings limit the possibility to obtain a smooth post-polished surface of the PVD coating. In contrast, post-polishing of the CVD coating does not suffer from intrinsic coating defects.

- The initial transfer of workmaterial to the tool rake face is significantly reduced by substrate pre-treatment and post-polishing the deposited coating. An improved surface finish was found to significantly reduce the crater wear rate of the cutting inserts investigated. This can most likely be explained by the reduced tendency to discrete chipping of coating fragments in the contact zone and the formation of a thin transfer layer composed of Al, Si, Ca, O with beneficial friction properties, both promoted by a smooth coating surface.

- Wear mechanisms of the coating used for turning stainless steel 316L seem to be controlled by superficial plastic deformation in combination with discrete chipping of coating fragments.
In turning stainless steel 316L with uncoated cemented carbide tools, the results strongly supports the hypothesis that dominant wear mechanisms of uncoated cemented carbide tools is related to tribochemical diffusion wear. It is believed that the observed crater wear of cemented carbide inserts at high cutting speeds is mainly the result of atomic diffusion of W and C into the chip sliding over the surface. The crater wear of the cemented carbides tends to increase with decreasing WC grain size. This is mainly due to a higher interfacial micro contact area attributed to the smaller WC grain size promoting diffusion wear. In the flank region a combination of mechanical wear (superficial plastic flow) and diffusion wear are the dominating wear mechanisms. The flank wear of the cemented carbides tends to decrease with decreasing WC grain size. This is mainly due to an increase in mechanical strength (increase in hardness) with decreasing WC grain size increasing the resistance to superficial plastic flow contributing to the wear on the flank wear land.

In turning of Ti6Al4V with uncoated cemented carbide, for combinations of low cutting speeds and feeds, crater and flank wear are controlled by an attrition wear mechanism. For combinations of medium to high cutting speeds and feeds, a diffusion wear mechanism was found to control the wear. Diffusion wear of cemented carbide during turning Ti6Al4V occurs through atomic diffusion into the transfer layer, resulting in a diffusion zone (layer). High resolution SEM, EDS and AES depth profile analysis reveal the formation of an approximately 100-150 nm thick carbon depleted WC-layer and TiC precipitation at the cemented carbide/Ti6Al4V interface.
The wear of a cutting tool is very complex. Severe conditions at the tool/work material sliding interface result in high mechanical and thermal stresses. In common for the wear of uncoated and coated cutting tools is that wear mechanisms act on a fine (nm-μm) scale. This indicates that knowledge about the microstructure of the coatings and the tool substrate material is of utmost importance. Depending on the dominant wear mechanisms, different microstructures could perform better or worse. In ideal cases, microstructures could be designed in order to reduce the effects from the dominant wear mechanisms.

- A conclusion from paper I is that a possible way towards reducing the wear rate of PCBN in machining could be to investigate the possibility of replacing cobalt with an alternative binder phase with lower chemical affinity to silicon oxide.

- In the never-ending research towards increasing the tool lifetime in the machining of difficult to machine materials new cemented carbide grades are continuously developed. When designing grades to prevent mechanical degradation, a high hardness/toughness is desirable to obtain high macro-mechanical strength of the cutting edge region. On the other hand, continuous wear mechanism occur on a very fine scale. When diffusion degradation dominates in a microstructure, decreasing the diffusion of elements across the sliding interface is the target. The present work shows an example in paper IV, in which larger WC grains and smaller binder phase pockets could be beneficial, if such a grade could be designed without dramatically reducing macroscopic toughness.

- The present work has shown an example of a wear mechanism in which cobalt in removed in a first stage during machining, and then binder pockets get filled by adhered material, which leads to a higher dissolution wear rate of WC grains. In this example, new cemented carbide grades with cobalt-free binder could increase wear resistance of cemented carbide.
Please note that any binder, significantly softer than WC, most probably will be removed from the surface during the initial stage. Thus, we should look for a more wear resistant binder without lowering the macroscopic strength of the cemented carbide microstructure.

- The question has been raised lately whether cobalt overall could be connected to health hazards. This has caused the machining industry to take an interest in proactively exploring the properties of alternative binders. Fe-based binders are used in metal cutting but their properties are still not suitable for machining steels, titanium and super alloys and still a need for further development is necessary [54]. Toller et al. [55] have tested turning inserts from cemented carbide with a nickel-iron binder and compared these with cobalt based reference inserts in steel turning. It is found that the life time of the cemented carbide with nickel-iron binder is only approximately 15% shorter than with the cobalt binder in the dry turning test, which motivates further studies with this alternative binder.

- According to the results presented in papers V and VI, in turning Ti6Al4V at medium and high cutting speeds/feeds a diffusion wear mechanism was found to control the wear of cemented carbide cutting tool. Modifying the surface of cemented carbide with “overstoichiometric” carbon on the surface which interact with the workpiece material before attacking the carbon of the WC grains is shown to be effective to increase the life time of the cutting tool [56]. PVD coating also gives some improvement in machining titanium alloys but the research of Sandvik Coromant team shows that the contribution of the PVD-coating to the tool performance is smaller than the contribution of the substrate [56]. There is also an incessant search for new grades of cemented carbide that can withstand the chemical breakdown by doping of the WC crystals [54, 57-59].

- Diffusion wear is a thermal activated process and decreasing temperature will reduce diffusion rate. Recently cryogenic coolant (liquid nitrogen) has been used in turning titanium alloy in order to reduce cutting temperature and increase the lifetime of the cutting tool [53]. However, to increase the knowledge on cryogenic machining of Ti6Al4V and applying the best coolant is a topic of interest for the future study.

- Application of high-speed machining is of interest of industry because of its favourable characteristics such as high productivity and better work quality. But machining at high cutting speed using cemented
carbide cutting tool is always challenging due to diffusion rate of cemented carbide at high temperature. In order to understand the diffusion mechanism different techniques is used nowadays. Diffusion couples, modelling, and experimental studies. Studying diffusion couples will increase the understanding of diffusion mechanism by studying the reaction between cemented carbide and workpiece material. But it should be reminded that static diffusion couple tests cannot fully represent the dynamics of a machining trial since in machining trial, the elements of the cutting tool diffuse into the chip which is continuously renewed. This renewal does not allow a saturation of the diffusing elements at the tool-chip interface. Also numerical study without parallel processing is of less interest and the numerical results should be checked with the experimental data, then the importance of well controlled experimental study using high technology surface analysis techniques is highlighted.

- The results presented in this thesis showed that the combination of surface characterisation techniques such as SEM, EDX, AES and ToF-SIM analysis can be used in order to obtain useful and complementary information regarding the surface and interface characteristics of worn surfaces of cutting tools. Electron microscopy and surface analysis techniques can be used to clarify the wear mechanisms on a very fine scale and be applied to the development work to improve microstructure of the cutting tools.

The tungsten carbide crystal structure is hexagonal close-packed with anisotropic properties. The possibility to evaluate the influence of crystallographic orientation of WC crystals on the deformation and wear mechanisms in machining using electron backscatter electron diffraction (EBSD) based techniques should also be performed.
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