On the Machinability of High Performance Tool Steels

NATALIA SANDBERG
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Abstract

The continuous development of hot forming tool steels has resulted in steels with improved mechanical properties. A change in alloying composition, primarily a decreased silicon content, makes them tougher and more wear resistant at elevated temperatures. However, it is at the expense of their machinability. The aim of this study is to explain the mechanisms behind this negative side effect.

Hot work tool steels of H13 type with different Si content were characterised mechanically, and evaluated analytically and by dedicated machining tests. Machining tests verified that materials with low Si content displayed reduced machinability due to their stronger tendency to adhere to the cutting edge. Three hypotheses were tested.

The first hypothesis, that the improved toughness of the low Si steels is the reason behind their relatively poor machinability, was rejected after machining tests with one low Si steel heat treated to the same relatively low toughness as conventional hot work tool steels.

The second hypothesis, that a change in oxidation properties, also associated with the change in Si composition, lies behind the reduced machinability was investigated by dedicated tests and evaluations. It was found that the oxide thickness increased with reduced Si content and that there was an enrichment of Cr at the oxide/steel interface. The differences in oxide thickness and the possible differences in oxidation properties may influence the machinability of the materials through their different abilities to adhere to the cutting edge.

The third hypothesis, that a high enough temperature to initiate phase transformation from ferrite to austenite is generated during machining of the tool steels, was also investigated. This may lead to a reduced machinability because higher austenite content is directly related to higher compressive stresses and higher cutting forces. This causes accelerated tool wear. This hypothesis was verified by ThermoCalc calculation of austenite content in the steels, which showed a good agreement with Gleeble compression tests and cutting force measurements.

This thesis confirms that a reduced Si content in conventional H13 steel improves the toughness, reduces the oxidation resistance and lowers the ferrite-to-austenite transformation temperature. The reduction in austenite temperature is probably the most important factor behind the reduced machinability.

Keywords: hot work tool steels, alloying composition, machinability

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urn:nbn:se:uu:diva-172427 (http://urn.kb.se/resolve?urn=urn:nbn:se:uu:diva-172427)
To my husband Odd Sandberg and
to the memory of my brother Alexander
List of Papers

This thesis is based on the following papers, which are referred to in the text by their Roman numerals.


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The author’s contribution

Paper I  Part of planning, writing and evaluation.

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

1 Background .................................................................................................................. 9

2 Introduction .................................................................................................................. 11
  2.1 Tool steels .............................................................................................................. 11
    2.1.1 Manufacturing ............................................................................................... 13
    2.1.2 Machinability ................................................................................................. 16
  2.2 Cutting tool materials ......................................................................................... 18
  2.3 Aim of this thesis ................................................................................................. 19

3 Material and characterisation .............................................................................. 20
  3.1 Tested tool steels ................................................................................................. 20
  3.2 Cutting tests ............................................................................................................ 22
  3.3 Characterisation methods .................................................................................... 23
    3.3.1 Mechanical properties .................................................................................. 23
    3.3.2 Thermal properties ....................................................................................... 25
    3.3.3 Oxidation properties .................................................................................... 26
    3.3.4 ThermoCalc simulations .............................................................................. 26
    3.3.5 Finite Element Modelling (FEM) simulations .............................................. 27
    3.3.6 Microscopy techniques ................................................................................. 28

4 Hypothesis .................................................................................................................. 31
  4.1 Toughness – Hypothesis 1 (Papers II and IV) ...................................................... 31
  4.2 Oxidation – Hypothesis 2 (Paper III) .................................................................. 31
  4.3 Austenitizing temperature – Hypothesis 3 (Papers IV and V) ......................... 32

5 Summary of results .................................................................................................... 33
  5.1 Machinability of steels with varying silicon content (Paper I) ................. 33
  5.2 Influence of toughness on machinability - Hypothesis 1 (Papers II and IV) ................................................................. 34
  5.3 Relation between oxidation properties and machinability – Hypothesis 2 (Paper III) ................................................................. 37
  5.4 Influence of austenitizing temperature – Hypothesis 3 (Papers IV and V) ................................................................. 41
  5.5 Thermal properties of H13 and H13M and their influence on the cutting temperature ................................................................. 43
Preface

The research associated with this PhD thesis has been carried out at Research and Development Department, Machining Section, at Uddeholms AB in Hagfors, and at the Tribomaterials Group, Ångström Laboratory, at Uppsala University. The financial and material support from the participating companies in the research project Böhler Uddeholm AG, Sandvik Tooling AB and Oerlikon Balzers Sandvik Coating AB is gratefully acknowledged. I am grateful to many people, who have inspired, encouraged, supervised me and have been helpful during the research and writing this thesis.

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Finally, I would like to thank my family for their great support.

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April, 2012
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1 Background

Tool steels constitute a very special group of alloys having different properties than other groups of steels. Tool steels are carbon, alloy or high-speed steels capable to be hardened and tempered and produced under strict processing. Tool steels are primarily used in dies and moulds for shaping, forming, blanking and cutting of other materials at ambient or elevated temperatures. Normally, tool steels are produced in electric furnaces. Special controlled heat treatment gives the steels their required properties. The first tool steels were made in the second half of 19th century [1].

The industrial demands on tool steel performance have increased because of the need to reduce costs [2, 3, 4]. Production of large components with thin wall thicknesses, complicated shapes and closer tolerances leads to more damage of the die surfaces, like heat checking and hot wear or even catastrophic gross cracking of the whole die. This has led to the development of hot work tool steels with higher toughness, higher strength and higher resistance to softening when operating at elevated temperature.

There are many ways to improve the mechanical properties of such materials. One way is to decrease the number of non-metallic inclusions and precipitated primary carbides in the steel, which could be achieved by a modification of the basic chemical composition of the steel. Carbide precipitation is decreased by reducing the silicon content in the steel. The combination of a low silicon content and a raised molybdenum content results in an increased toughness and higher strength at elevated temperatures. However, these die steels with modified chemical composition have been considered difficult to machine in both the soft annealed and the hardened and tempered state, although the effect is greatest in hardened material, when comparing with conventional AISI H11/H13 type of grades.

Good steel machinability is defined as a combination of low power consumption, low tool wear and good surface finish. Tool life is one important factor in metal cutting from an economical point of view. Cutting tool flank wear is influenced by the interaction between the cutting tool and work piece. New hot work tool steels with improved mechanical properties generate relatively high cutting forces and edge temperatures, which, in turn, promote sticking of tool steel to the cutting edge and an associated built up edge formation. This results in shorter cutting tool life. The high machining cost is in most cases negligible in relation to the increased performance in die life, but should still be kept as low as possible.
2 Introduction

2.1 Tool steels

Many grades and compositions of tool steels are known. Most tool steels can be classified according to the American Iron and Steel Institute (AISI). According to the classification the tool steels are divided in the following groups:

- Water-hardening tool steels
- Shock-resisting tool steels
- Oil-hardening cold work tool steels
- Air-hardening, medium alloyed cold work tool steels
- High-carbon, high-chromium cold work tool steels
- Mould steels
- Hot work tool steels, alloyed with chromium, tungsten, and/or molybdenum
- Tungsten high-speed tool steels
- Molybdenum high-speed tool steels

One of the biggest and most important groups of the tool steels is the hot work group.

Hot work tool steels are divided into three types: chromium hot work tool steels H10-H14 and H19, tungsten hot work tool steels H21-H26 and molybdenum hot work tool steels H42, according their principal alloying elements [1]. The hot work tool steels are usually alloyed to low or medium carbon content (0.35-0.45 wt.% ) to provide high toughness, and additions of carbide forming elements such as chromium, tungsten, molybdenum and vanadium to provide high temper resistance and hot hardness.

The chromium hot work tool steels are relatively low alloyed with Cr content of about 3-5 wt.% in addition to ≤ 1 wt.% V and ≤ 3 wt.% Mo. The relatively low alloy content provides toughness and hardness about 40-55 HRC in the heat treated condition. The alloying elements also contribute to give a high hardenability by securing a transformation to desired martensitic microstructure during quenching. These tool steel grades must have good resistance to high temperature impact loading, to softening during high temperature exposure, and to thermal fatigue. The medium carbon content provides toughness by lowered carbon concentration in martensite and by limiting size of alloy carbide particles. Fine dispersed chromium and vanadium carbides precipitate when tempering at high temperature, which provides
good high temperature strength. These carbides are very stable and coarsen very slowly in service. The chromium hot work steels are widely used for extrusion of aluminium, die casting of light metals (Al, Mg) and steel forging applications [1].

The less common tungsten hot work tool steels contain high contents of tungsten, about 9-18 wt.%. The high alloying content provides enhanced thermal stability and much higher resistance to softening at elevated temperatures compared with the chromium hot work tool steels, but also makes them more prone to brittleness. Addition of tungsten improves resistance to softening, but at the same time it produces a larger volume fraction of stable alloy carbides in the microstructure, which results in reduced toughness. Tungsten hot work tool steels are used for extrusion dies for brass and bronze and punching of hot material [1].

The principal alloying elements in molybdenum hot work tool steels are molybdenum, tungsten, chromium and vanadium. This type of hot work tool steel has similar properties and applications as tungsten hot work tool steels.

The increasing demands on die cast products has lead to the development of hot work die steels with higher strength and toughness, improved polishability and weldability, and high dimension stability.

One way to improve the mechanical properties of the hot work tool steels is to modify their chemical composition. Small difference in composition has a significant influence on the properties of the material.

The volume fraction of carbides in the steel depends on the carbon content. The hardness of martensite as well as the ultimate tensile and yield strengths increase with increased carbon content [5]. Addition of carbon improves strength and hardenability of the steels, but reduces their toughness.

Chromium is a carbide forming alloying element. Increasing the chromium content results in an increased hardenability and resistance to corrosion and oxidation, and contributes to some extent to the high temperature strength and abrasion resistance in high-carbon steels [5].

Molybdenum is also a strong carbide forming element, and is a more effective carbide forming element than chromium. Increased molybdenum content results in an increased resistance to thermal softening. Addition of molybdenum also increases the strength, the toughness and the hardenability of the hot work tool steels [5].

Tungsten is an even stronger carbide forming element, and addition of this element contributes to high creep strength in high-temperature alloys. Tungsten gives similar property as molybdenum, but needs to be present in large quantity.

Manganese is an alloying element that mainly contributes to hardenability at a moderate cost. Increasing the manganese content in combination with sulphur also improves the machinability due to formation of manganese sulphide. Manganese in the steels improves their strength in the ferritic phase [5].
Silicon is not a carbide forming element, but it enhances the carbon activity, and thus promotes formation of unwanted primary carbides. Increasing the silicon content contributes to a moderate hardenability, increases the strength of quenched and tempered steels, and slightly increases the strength of retained austenite. Silicon addition also improves oxidation resistance and is generally used as a deoxidiser [5]. S. Taniguchi et al. [6] found that the oxide layer on the surface of the material became thinner with increasing Si content due to the formation of Fe$_2$SiO$_4$ grains at the interface between the oxide scale and the substrate. This layer acts as an obstacle for iron diffusion from the substrate to the scale, and formation of FeO is prevented. G. Bamba et al. [7] showed that the presence of silicon in Fe-Cr steels reduces oxidation rates by formation of internal amorphous silica, and prevents chromium supply to the oxide scale.

Sulphur in steel has a negative effect on mechanical properties, but small amounts of this element improve machinability through the formation of soft MnS particles. These particles are nuclei for cracks, which facilitates separation of the chip from the cutting tool, and results in thinner chips, lower cutting forces, and lower power consumption [8].

2.1.1 Manufacturing

2.1.1.1 Steelmaking

A wide variety of manufacturing equipment is used for production of tool steels and may also be followed by subsequent processing as electroslag remelting (ESR) or vacuum arc remelting (VAR) for enhanced properties. The different steps of a conventional ingot casting route followed by electroslag remelting are shown in Fig. 1. Primary melting is performed in an electric arc furnace (EAF) where scrap and ferroalloys are melted with little or no refining. Special care must be taken to avoid scrap contamination from elements such as nickel, cobalt and copper, which can not be oxidised out of the melt. In a second stage, the hot metal from the EAF is transferred to a ladle or converter vessel where the majority of refining is performed [1]. Operations like argon oxygen decarburization (AOD), vacuum oxygen decarburization (VOD) and ladle furnaces are secondary refining methods. The treatment in a ladle furnace station starts with deslagging to remove the oxygen rich EAF-slag. Deoxidation elements like aluminium is subsequently added to reduce soluble oxygen in the melt. The deoxidation is followed by addition of various alloys, which depends on the specification of the steel grade. At the same time, a new top slag is introduced to facilitate desulphurisation, capture of non-metallic inclusions from the steel melt and to protect the steel melt from reoxidation. When the alloys have been dissolved, the appropriate steel composition has been attained and the temperature is in the right interval then the ladle is transferred to a degassing station. There argon
gas is purged into the melt for reduction of nitrogen and hydrogen. The stirring promotes growth and separation of inclusions from the steel into the top slag. A combined gas and induction stirring is an effective way to produce clean steel. When the desired chemistry has been achieved and the proper temperature reached, the ladle is transferred to the casting station. Except for low alloyed tool steels, ingot casting is performed as bottom pouring ingots moulds. Otherwise, continuous casting may be an alternative for a cost reduction. Following casting, ingots are normally annealed to prevent cracking or stripped and heated directly for forging or rolling.

Electroslag remelting (ESR) is a process to obtain steel of high cleanliness with improved mechanical properties. The end of an ingot cast electrode is submerged to molten slag in a water-cooled copper mould and high current is applied. As the electrode consumes, droplets of molten metal fall through the slag, where it is refined and reduced in sulphur and oxygen. Finally, it is deposited at the melt pool bottom to participate in ingot build up. The ESR ingot has a uniform microstructure with low porosity and with reduced center segregation. The ESR process operates at low rates, typical 500-1000
kg/h melt rate, and the rather high production cost makes it applicable to tool steels when improved yield, high cleanliness, low sulphur content and a refined structure are of significant importance.

2.1.1.2 Heat treatment

Heat treatment of tool steels is a critical process, because it influences the final properties of the materials. In soft annealed tool steels, most of the alloying elements are bound as carbides. When the steel is heated for hardening the fundamental idea is to dissolve the carbides to such a degree that the matrix acquires an alloying content that gives the hardening effect without becoming coarse-grained and brittle. This means that the matrix is alloyed with carbon and carbide forming elements. A transformation from ferrite to austenite occurs. If the steel is quenched rapidly in the hardening process, the carbon atoms do not have time to reposition themselves to allow reforming of ferrite from austenite. Instead they are fixed in positions that results in high microstresses and as a consequence hardness is increased. This structure is called martensite. Untransformed austenite, called retained austenite, is always present in the quenched structure. The amount of austenite increases with the amount of alloying elements, higher hardening temperature and longer holding time. After quenching, the microstructure of the steel contains martensite, retained austenite and carbides. A tempering should be performed immediately after the hardening.

During hardening, protection from oxidation and decarburisation is of great importance. The heating is performed in vacuum furnaces or furnaces with a protective gas atmosphere. Heating and cooling rates are controlled. When the steel reaches the hardening temperature through its entire thickness, it is normally held at this temperature for 30 min, cp. Fig. 2. Quenching rates have a significant influence on the steel microstructure. Fast quenching rates result in the best microstructure and tool performance, but also the risk of tool cracking. To minimize this risk slower quenching rates should be used to reduce the temperature difference between the surface and the core. Martensite formation leads to an increase in volume and stresses in the material, and the quenching should be interrupted before room temperature is reached. Too slow quenching rates can lead to undesirable transformations in the microstructure, which may result in deterioration of tool performance.

A tempering should follow immediately after the quenching to reduce the risk of cracking. Two temperings are normally used, see Fig. 2. After the first tempering, the microstructure consists of tempered martensite, newly formed martensite, some retained austenite and carbides. Precipitated secondary carbides and newly formed martensite can increase the hardness during tempering, i.e. secondary hardening occurs. The second tempering has a significant effect on the secondary martensite formed after the first tempering. The tempering temperature depends on factors such as hardness, toughness and dimensional changes. Higher tempering temperature results in
lower hardness. Normally, tempering temperatures around 550 to 620ºC are used. The holding time for each tempering is 2 hours.

![Schematic diagram of hardening and tempering of tool steels.](image)

2.1.2 Machinability

Good steel machinability is defined as a combination of low power consumption, low tool wear and good surface finish [9]. Tool life is one important factor in metal cutting from an economical point of view and has been used as machinability criterion in this thesis. The overall goal of machining is to remove a certain volume of material at highest possible rate with as low cost as possible. The machining process is associated with severe deformation, which results in high energy consumption, high local temperatures and wear of the cutting tool [8].

Some of the new high performance tool steels are difficult to machine. Machining dominates the cost in tool production, cp. Fig. 3. Hence, enhanced machinability would reduce the cost of machining operations through less cutting tool consumption, power consumption and operation time.

![Cost distribution for manufacturing of a die casting tool.](image)

Machinability of tool steels is influenced by many factors, such as chemical composition, microstructure, inclusions and thermo-mechanical properties. The optimum microstructure for machining is a uniform distribution of spheroidized carbides in a soft annealed ferritic structure with as low hard-
ness as possible. The machinability of martensitic hot work steels is mainly influenced by the amount of non-metallic inclusions like manganese sulphides, and the hardness of the steel.

The power consumed in machining is mostly converted into heat near the cutting edge. The heat generated at the tool-work piece interface has a significant influence on the cutting tool and surface quality. High temperatures at the interface lead to thermal softening of the cutting tool, diffusion of chemical elements between tool and work material, loss of dimensional accuracy and deterioration of surface quality by inducing residual stresses and surface hardening.

Cutting tool life is an important aspect of machinability and directly influences the quality of the machined surface. The wear and fracture of the cutting tool results in poor surface finish and loss of accuracy. In practice a longer cutting tool life is synonymous with a higher productivity in forming tool production [10].

Hot work tool steels are normally delivered in the soft annealed condition. After machining, the die must be heat treated in order to achieve an optimum hot yield strength, temper resistance, toughness and ductility. This thesis is focused on machining of hardened and tempered hot work tool steels.

When a die casting die is hardened and tempered some distortion normally occurs. By leaving some machining allowance on the die prior to hardening it is possible to adjust the die to the correct dimensions after the hardening and tempering by, for example, grinding or hard machining.

2.1.2.1 Influence of mechanical and thermal properties on machinability

Ductility has a strong influence on machinability. Ductility is a measure of how much deformation a material can sustain before it fractures. Machining is a process of transformation of metal into chips. Materials with low ductility would exhibit better machinability. Ductility also has a strong influence on the cutting temperature. The plastic deformation during cutting generates a large amount of heat, which leads to a temperature rise in the primary deformation zone. Materials with high ductility show a tendency to higher adhesion to the cutting tool and an accelerated wear of the tool [9, 11]. Ductility and hardness have an inverse relationship. Hard materials are less ductile, which would make them comparatively easy to machine, but if a material is too hard its low ductility will be offset by its hardness. It is well known, that a higher hardness often results in lower machinability under identical conditions. In respect to machinability an ideal material should be not too hard and not too ductile.

High toughness is an essential material characteristic for hot work tool steels. A brittle material absorbs less energy than a tough material before it fractures [12]. The optimum combination of strength and ductility would result in maximum toughness. The higher the toughness of the material, the
higher is the force required to break it down. Therefore, materials with high toughness display reduced machinability.

Fracture toughness is another important property of the tool steel. It describes the ability of a material containing a crack to resist fracture. The stress-intensity factor (K) is a measure of the amount of stress required to propagate a pre-existing flaw [12]. Therefore, materials with low fracture toughness would be easier to machine.

The flow stress in compression is another parameter that has a strong influence on the cutting force. The higher the flow stress, the more difficult it is to deform the material and therefore a higher cutting force is required.

Thermal conductivity has also a bearing on machinability. Materials that conduct heat well facilitate the dissipation of heat away from the area that is being machined. Therefore, high thermal conductivity generally aids the machining process.

A low thermal conductivity in the steel creates higher temperatures in the cutting region, which results in higher cutting edge temperatures and increased wear of the cutting tool [13, 14]. Machining of steels with low thermal conductivity demands cutting tools with an ability to withstand high temperatures. Higher thermal conductivity in combination with faster heat transport reduces wear of the cutting tool.

2.2 Cutting tool materials

The development of cutting tool materials has followed the pressure of technological change and economic competition. The simplest concept of the requirements to the cutting tool is high hardness, high resistance to heat, and high toughness to withstand impact without fracture. The two most common materials for industrial cutting tools are high-speed steels and cemented carbides [8].

Cemented carbides (CC) are metal based composites that include carbides, nitrides and borides. WC grains in a Co matrix are the most widely used type of CC in metal cutting. They are produced by powder metallurgy, which makes it possible to control both the composition and the carbide grain size [8]. The CCs used in cutting tools normally contain at least 80 vol.% carbide phase. Carbides are metallic in character. They have good electrical and thermal conductivities, and very low ability to deform plastically without fracture. The hardness of carbides drops with increasing temperature, but they are always much harder than steel. The Co matrix of the CC inserts gives a high toughness and a high hot hardness.

In this thesis uncoated and PVD coated WC/Co-based CC inserts were used. The CC grade was H10F with the nominal chemical composition (wt.%) 10.0 Co, 89.5 WC and 0.5 Cr3C2. This CC was chosen due to the high melting point at 2750°C and high hardness of 2100 HV.
TiN coated CCs were used to study the evolution of wear of the cutting tool at different stages. The thickness of the TiN coating was about 4 µm. TiAlN coated CCs were used in face milling. The thickness of the multilayered coating was about 5 µm on the rake face and 2 µm on the clearance face. Uncoated CCs were used for a deeper study of the mechanisms of accelerated wear of the tool after coating removal.

2.3 Aim of this thesis

The work presented here is an analysis of the influence of the relative chemical composition on the machinability of two hot work tool steels produced by Uddeholms AB; one reference steel Orvar Supreme (H13), which is relatively easy to machine, and one modified H13 version (H13M), which is relatively difficult to machine. There is a significant difference in silicon content between these steels (see Table 1). Consequently, explaining how silicon influences the machinability of these two steels is the principal aim of this thesis.
3 Material and characterisation

3.1 Tested tool steels

Orvar Supreme (H13) and modified Dievar (H13M) are chromium-molybdenum-vanadium alloyed hot work tool steels.

H13 is characterised by a high resistance to thermal shock and thermal fatigue, high temperature strength, high toughness and ductility, good machinability and polishability, good hardening properties and dimensional stability during hardening. It is used in applications such as tools for die casting, extrusion and hot pressing, moulds for plastics, and in cold punching and hot shearing, as shrink rings and wear resistance part.

H13M is a high performance hot work tool steel, which is characterised by a higher toughness and ductility than H13, and also better temper resistance, high temperature strength, hardenability and dimensional stability throughout heat treatment and coating deposition. H13M is used for hot work applications such as tools for die casting, extrusion and hot forging. Compared to H13, H13M generally has better mechanical properties in hot applications.

The main difference in alloying composition between the two steels is that H13M contains less Si and V, but higher Mo content than H13 (see Table 1). Note that the Si content of H13M in this thesis has a lower value than in the specification of the commercial H13M (0.15-0.25 wt.%). This is to exaggerate any impact of silicon on the machinability. The microstructure of both steels is tempered martensite, see Fig. 4. The average grain size is 17 µm in H13 and 30 µm in H13M.

Both steels were produced by conventional ingot casting followed by electroslag remelting and forging, see Fig. 1. They contain very low amount of non-metallic inclusions.

The chemical composition of all materials tested in this thesis is presented in Table 1.
Table 1: Nominal chemical composition (wt.%) as measured with XRF, hardness and Charpy-V impact energy at 20°C of the tested materials

<table>
<thead>
<tr>
<th>Material</th>
<th>HRC</th>
<th>Energy, J</th>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>Cr</th>
<th>Mo</th>
<th>V</th>
<th>Ni</th>
<th>Co</th>
</tr>
</thead>
<tbody>
<tr>
<td>H13</td>
<td>45</td>
<td>22</td>
<td>0.39</td>
<td>1.03</td>
<td>0.38</td>
<td>5.18</td>
<td>1.45</td>
<td>0.89</td>
<td>0.08</td>
<td></td>
</tr>
<tr>
<td>H13M</td>
<td>45</td>
<td>28</td>
<td>0.35</td>
<td>0.06</td>
<td>0.48</td>
<td>5.05</td>
<td>2.34</td>
<td>0.55</td>
<td>0.10</td>
<td></td>
</tr>
<tr>
<td>H13M-L</td>
<td>45</td>
<td>18</td>
<td>0.35</td>
<td>0.06</td>
<td>0.48</td>
<td>5.05</td>
<td>2.34</td>
<td>0.55</td>
<td>0.10</td>
<td></td>
</tr>
<tr>
<td>QRO90</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Supreme</td>
<td>46</td>
<td>10</td>
<td>0.39</td>
<td>0.26</td>
<td>0.75</td>
<td>2.64</td>
<td>2.29</td>
<td>0.82</td>
<td>0.08</td>
<td></td>
</tr>
<tr>
<td>Stavax ESR</td>
<td>50</td>
<td>7</td>
<td>0.39</td>
<td>0.91</td>
<td>0.57</td>
<td>13.3</td>
<td>0.28</td>
<td>0.29</td>
<td>0.14</td>
<td></td>
</tr>
<tr>
<td>Stavax</td>
<td>48</td>
<td>10</td>
<td>0.24</td>
<td>0.18</td>
<td>0.62</td>
<td>13.3</td>
<td>0.35</td>
<td>0.34</td>
<td>1.35</td>
<td></td>
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<tr>
<td>Sim. comp.*</td>
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<td></td>
<td></td>
<td></td>
<td></td>
<td>5.0</td>
<td></td>
</tr>
</tbody>
</table>

* The simulated compositions represent four alloys evaluated by ThermoCalc. Each of them has a composition identical to H13M, but with a change in one of the elements Mn, Cr, Mo and Co at time.

Fig. 4: The microstructure of a) H13M and b) H13.

H13 and H13M were hardened and tempered to a hardness of 45 HRC. Special conditions for heat treatment were used for H13M-L to achieve the same toughness level as that of H13. Heat treatment conditions are presented in Table 2 (Paper IV).

Table 2: The heat treatment of H13 and H13M

<table>
<thead>
<tr>
<th>Material</th>
<th>Pre-heating, °C</th>
<th>Austenitizing, °C</th>
<th>Cooling</th>
<th>Tempering, °C</th>
</tr>
</thead>
<tbody>
<tr>
<td>H13, H13M</td>
<td>First -650, 30 min Second- 850, 20 min</td>
<td>1020, 40 min</td>
<td>Polymer</td>
<td>610, 2 x 2 h</td>
</tr>
<tr>
<td>H13M-L</td>
<td>First -650, 30 min Second- 850, 20 min</td>
<td>1030, 150 min</td>
<td>Polymer</td>
<td>590, 4 h 600, 2 x 4 h</td>
</tr>
</tbody>
</table>

23
3.2 Cutting tests

The machinability of the forming tool steels was evaluated by using interrupted turning and milling.

The cutting was performed in a turning lathe with special geometry of the work piece to simulate interrupted cutting, see Fig. 5a. This method combines some of the conditions typical of both milling and turning operations. Bars designed for interrupted orthogonal cutting, see Fig. 5b, were used for cutting force measurements. A Kistler force measurement dynamometer platform was used in the test.

![Image](image.png)

Fig. 5: Geometry of the work piece used in machining tests to evaluate cutting tool wear (a) and for cutting force measurement in interrupted cutting (b).

To avoid any influence of tool wear in the cutting force measurements, new edges were used in each experiment, and the forces were measured during three consecutive tool/work engagements (engagement no 5, 6 and 7). Chips and cutting inserts from the force measurements were collected and studied by scanning electron microscopy (SEM) and energy dispersive spectroscopy (EDS).

The interrupted turning tests were performed dry and with the same cutting parameters for both steels despite their difference in machinability. Normally a higher cutting speed is used for H13.

Wear of the TiN coated cemented carbides used in interrupted turning under the same cutting conditions in H13 and H13M was also studied by using 3D white light interferometry.

Single tooth face milling with TiAlN coated carbide inserts was used to confirm the results obtained from the interrupted turning and for study of the wear evolution of the cutting tool at different stages and to evaluate the cutting tool life. There are many differences between these operations, but the principal difference between interrupted turning and face milling is the chip
formation. In the interrupted turning the chip has a constant thickness while it is thicker at the entrance and thinner at the exit in down milling and the opposite in up milling. The milled area was 45 x 80 mm. Pre-milled work pieces were used for achieving full tool engagement (full chip length) and to avoid the influence from tool wear in the cutting force measurement. The geometry of the pre-milled work piece used in milling test is shown in Fig. 6. New cutting edges were used in each experiment. Milling tests were performed without coolant.

3.3 Characterisation methods

3.3.1 Mechanical properties

Since the mechanical and thermal properties of the material are important factors in machining, they must be properly characterised. The strength, ductility and toughness can be evaluated by using different methods [15].

3.3.1.1 Tensile test

To evaluate the ductility of the materials, tensile tests were performed at room temperature according to Standard EN ISO 6892-1:2009 [16] and at elevated temperatures (200-650°C) according to EN 10 002-5 [17]. The material testing machine used was a floor test Zwick Z250/SW5A with a maximum test load of 250 kN and designed for quasi-static loading. The test samples of type 10C100 were taken from the longitudinal direction of the test material bars. For each material and test condition three test samples were made and average value was reported.

3.3.1.2 Gleeble compression test

The flow stress in compression is a parameter that has a strong influence on the cutting force. Gleeble compression tests were used to compare flow stress between the investigated materials. They were performed in a Gleeble-3800 Hydro Wedge™ system especially adapted for cylinder compression test.
testing [18]. The Gleeble-3800 mechanical system is a complete, fully integrated hydraulic servo-system capable of exerting up to 20 tons of static force in compression or 10 tons in tension. Displacement rates up to 2000 mm/sec can be achieved. The action cylinder was controlled by a staged servo-valve. Test rods with diameter of 10 mm and length of 12 mm were used. In this thesis the test temperature ranged from 650 to 1000ºC with intervals of 50ºC, and the strain rates were fixed to 1, 10 and 100 s\(^{-1}\). From 810 to 880ºC intervals of 10ºC and a strain rate of 10 s\(^{-1}\) were used. The small temperature intervals in the temperature range from 810 to 880ºC were because the ferrite-to-austenite transformation.

3.3.1.3 Charpy-V impact test
In order to evaluate the toughness of the materials impact energy tests were performed at ambient and elevated temperatures (20-600ºC). A Roell Am- sler RKP150 machine with nominal energy of 150 J and release angle of 150º was used in accordance with ASTM E23-07 [19]. To ensure the validity of the test values, three tests were performed at each temperature. The test specimens were taken from the centre of cutting test bars in longitudinal direction, i.e. parallel with the bar axes.

3.3.1.4 Fracture toughness test \(K_{JC}\)
The \(K_{JC}\) value can be used as an estimate of fracture toughness near the initiation of slow stable crack growth for metallic materials [20]. The tests were performed in accordance with ASTM E1820-2009A [21]. The material testing machine used was a floor test MTS with a maximum test load of 100 kN and the fracture toughness tests were performed at elevated temperatures from 200 up to 600ºC.

3.3.1.5 Cutting tool wear measurement
The flank wear width of the cutting tools was measured using light optical microscopy (LOM). A cutting tool can wear in different modes, for instance by edge rounding, crater wear or flank wear, see Fig. 7. The most critical mode is flank wear. In these studies an average flank wear of 0.3 mm was used as the worn out criterion when interrupted turning. An average flank wear of 0.2 mm was used when single tooth face milling in hot work tool steels, and 0.3 mm in stainless plastic mould steels.
3.3.2 Thermal properties

Thermo-physical properties of the materials, i.e. diffusivity, heat capacity and thermal conductivity, play an important role when machining due to their influence on the cutting temperature [9, 22]. These properties can be evaluated with help of a thermal analysis apparatus and a dilatometer. Laser Flash LFA 457 was used for thermal conductivity measurement. The measurements are carried out in Argon gas and can be performed at room and elevated temperatures. Two references are usually used in the experiments, one of them is an Fe reference for temperatures up to 700°C, and the other is Inconel600 for temperature up to about 1100°C. The sample holder was made in SiC. An unfocused laser pulse heats the front face of the sample, which results in a temperature increase as a function of time on the back face of the sample. The samples are covered by a thin layer of graphite spray on the both sides to avoid reflection of the laser beam. The equipment gives the diffusivity \( D \), i.e. the measure of how fast the material transmit heat, according to the equation

\[
D = 0.1388l^2/t_{1/2}
\]

where \( l \) is the thickness of the sample and \( t_{1/2} \) is the time after which the temperature on the back face of the sample has reached half of its maximum value [23].

With help of data for length extension with temperature from dilatometer and density \( (\rho) \) measured at room temperature, the density at elevated temperature can be measured. Specific heat \( (C_p) \) can be measured as a function of temperature. Therefore the thermal conductivity \( (\lambda) \) at specific temperatures can be calculated according to the equation

\[
\lambda(T) = D(T) \cdot \rho(T) \cdot C_p(T)
\]
The tested samples had a diameter of 12.6 mm and thickness 2.5 mm. As a reference, Inconel600 sample with the same dimensions as the tests samples was used. The measurements were performed from 25°C up to 1000°C, i.e. when transformation ferrite-to-austenite occurs.

The measured \( C_p \) values were used in the simulation of cutting temperature in § 3.3.5.

### 3.3.3 Oxidation properties

Differences in the oxidation properties between steels may be a reason for the differences in machinability. If a change in the alloying elements results in a composition of the material surface that approaches stainless steel, then this steel becomes difficult to machine, partly because of a high sticking ability to the cutting edge [24].

For characterisation of oxidation properties of the material two techniques with nanometer depth sensitivity were applied; glow discharge optical emission spectrometry (GDOES) and X-ray photo electron spectroscopy (XPS). These techniques record elemental depth profiles of surface layers. The analysed area in GDOES is 0.5 mm and layers from 1 nm to 100 µm can be analysed. The technique permits rapid analysis and very good detection of light elements like H, C, O, N etc.

X-ray photoelectron spectroscopy (XPS), also known as ESCA, Electron Spectroscopy for Chemical Analysis, is a quantitative spectroscopic technique for all elements except H and He.

XPS spectra are obtained by irradiating a material with a beam of monochromatic X-rays while simultaneously measuring kinetic energy and number of electrons that escape from the surface. The information depth is typically 1-10 nm, and compositional profiles can be recorded down to about 1µm depth. XPS requires ultra-high vacuum conditions, while GDOES performed in air. For the GDOES analysis a Leco GDS850A, calibrated with certified reference materials, was used.

### 3.3.4 ThermoCalc simulations

Since austenite has totally different mechanical and thermal properties than ferrite, and therefore displays different machinability, the difference in transformation temperature from ferrite to austenite (A1) between the investigated hot work tool steels was simulated. For this purpose a powerful software for thermodynamic and phase diagram calculations in multicomponent system, ThermoCalc, has been used [25].

ThermoCalc calculations were performed to estimate the vol.% of austenite phase in the 800-950°C temperature interval. The effect of small changes in alloying of H13M on the austenite volume at given temperatures was also calculated. These calculations were performed for each element separately.
3.3.5 Finite Element Modelling (FEM) simulations

In order to evaluate the difference in cutting temperature that can occur due to the difference in thermal properties of the two work materials represented by the two forming tools studied in this thesis, Finite Element Modelling (FEM) was used. The general FE-code ABAQUS™ Standard was used for the simulation. Steady state temperature distributions were computed using a fixed two-dimensional mesh with an Eulerian description of the material flow through the elements. The orthogonal cutting conditions are presented in Table 3.

<table>
<thead>
<tr>
<th>Parameter</th>
<th>Value</th>
</tr>
</thead>
<tbody>
<tr>
<td>Initial temperature, ( T_0 )</td>
<td>20°C</td>
</tr>
<tr>
<td>Rake angle, ( \gamma )</td>
<td>0°</td>
</tr>
<tr>
<td>Feed, ( s )</td>
<td>0.2 mm/rev.</td>
</tr>
<tr>
<td>Chip thickness, ( t )</td>
<td>0.34 mm</td>
</tr>
<tr>
<td>Cutting speed, ( v )</td>
<td>110 m/min</td>
</tr>
<tr>
<td>Length of friction heat zone</td>
<td>0.68 mm</td>
</tr>
<tr>
<td>Friction heat, ( q )</td>
<td>2.5·10^8 Wm(^{-2})</td>
</tr>
<tr>
<td>Thickness of volume heat zone (primary deformation zone)</td>
<td>1 ( \mu )m</td>
</tr>
<tr>
<td>Deformation heat, ( Q )</td>
<td>1.1·10^{15} Wm(^{-3})</td>
</tr>
<tr>
<td>Heat transfer coefficient to air, ( \alpha )</td>
<td>40 Wm(^{-2})K(^{-1})</td>
</tr>
<tr>
<td>Air temperature, ( T_{\text{air}} )</td>
<td>20°C</td>
</tr>
</tbody>
</table>

It was assumed that the same amount of heat was generated from friction and deformation for both materials. Accordingly, the result reflects only the influence of the differences in thermal properties of the work materials on the cutting temperature [26]. The model includes the work material with two heat sources, an internal heat source at the primary deformation zone and a surface heat source from friction at the rake interface. The material flow and cutting parameters are shown in Fig. 8.

To reduce the impact from the boundaries of the model on the temperature close to the chip, the mesh included not only the chip but also a 50 mm x 50 mm section representing undeformed material. At boundaries the temperature was fixed equal to start temperature \( T_0 \). A boundary condition representing convection heat transfer was set at the external surfaces. The surface heat transfer was described by the equation

\[
q = \alpha \cdot (T - T_{\text{air}}) \quad \text{eq. (3)}
\]

The primary deformation zone was modelled as a volume heat source generating a power \( Q \) per volume.
3.3.6 Microscopy techniques

Microstructure of the material and wear of cutting tools can be studied with different microscopy techniques such as light optical microscopy (LOM), scanning electron microscopy (SEM) and transmission electron microscopy (TEM).

3.3.6.1 LOM and SEM

Light optical microscopy (LOM) is widely used for study of the general microstructure of materials. The spatial resolution of LOM is only about 0.5 µm and its depth of focus is very small, which makes it limited to features such as grain size and coarse particles from polished and etched cross-sections. Deeper studies of the microstructure require higher resolution and greater depth of field. Here scanning electron microscopy (SEM) is widely used. SEM images are produced by scanning a sample with a high-energy beam of electrons. The interaction between electrons and atoms in the sample surface produces signals that are used for both topographic imaging and characterisation of elemental composition. Secondary electrons give the best information about the surface topography. The elemental composition is analysed by X-rays. SEM operates in vacuum atmosphere, and therefore it requires not only electrically conductive but also vacuum tolerant samples.

SEM FEI Quanta 600F operated at 30 kV and equipped with an Oxford Instruments EDS system, Inca Feature software and motorized specimen table used on coated and uncoated cutting inserts to reveal tool wear and any adhered work material. SEM was also performed on the chips collected from the force measurement tests to reveal any adhered tool material.
3.3.6.2 TEM and FIB

When studies of the microstructure, crystallographic and electronic structure and chemical identity in the highest resolution are required transmission electron microscopy (TEM) is normally used. TEM provides information with a resolution, of the crystallographic structure of the order of a few Ångströms, and on the chemical composition within a few nm. In TEM a beam of electrons is transmitted through an ultra thin specimen, interacts with the specimen as it passes through, and forms an image.

TEM sample preparation is quite complicated. A modern instrument that combines a focused ion beam (FIB) with a SEM (FIB-SEM) was used to prepare the TEM specimens studied in this thesis. The FIB-SEM uses a finely focused beam of gallium ions. The ion beam can be operated at low beam currents for imaging or high beam current for site specific sputtering or milling. TEM sample are produced by using high beam current for milling.

Two uncoated inserts used to cut in the investigated steels were selected for high resolution studies using transmission electron microscopy (TEM; FEI F30ST TEM operated at 300 kV). The cross-section TEM samples were produced by means of a focused ion beam (FIB; FEI Strata DB 235 dual beam station operated with 30 kV Ga ions) technique.

3.3.6.3 3D white light interferometry

Evaluation of cutting tool wear can be done by studying topography of the tool surface. For this purpose a 3D white light optical profiler was used in this thesis. It offers fast, non-contact, high-precision 3D metrology of surface features with the combination of good resolution and repeatability. The method is based on white-light interferometry, a traditional technique in which a pattern of bright and dark lines (fringes) result from an optical path difference between a reference and a sample beam. Incoming light is split inside an interferometer. One beam goes to an internal reference surface and the other to the sample. After reflection, the beams recombine inside the interferometer, undergo constructive and destructive interference and produce a light and dark fringe pattern that is transformed into a topographical image. Optical profilers are able to measure sub-nanometer surface roughness with a lateral resolution down to about 1 µm. There is a wide range of optical profilers.

Evolution of wear of TiN coated cutting inserts was examined using a Wyko NT3300 3D white light interferometer at magnification 2.53 times (measured area 2.4 x 1.9 mm, sampling interval 3.32 µm) and cutting time 10 s, 1 min, 3 min and 5 min.
3.3.6.4 XRF

X-ray fluorescence (XRF) is widely used for elemental and chemical analysis, particularly in the investigation of metals. The method is based on the emission of characteristic "secondary" (fluorescent) X-Rays from a material that has been excited by bombarding with high-energy X-rays or gamma rays. For measurement of nominal chemical composition of the materials investigated in this thesis, ARL 9800XP with semi-quantitative analysis was used.
4 Hypothesis

The machining tests reported in Paper I indicated that a low silicon content is detrimental to the cutting tool life. Two hot work tool steels, H13 with about 1.03 wt.% Si and H13M with about 0.06 wt.% Si were selected for a thorough study of this strong influence of Si on their machinability.

Three different hypotheses have been suggested to explain this influence. They are briefly presented below in chronological order.

4.1 Toughness – Hypothesis 1 (Papers II and IV)

The toughness of the work material is one of many important parameters in machining. The presence of Si in steel increases the carbon activity. Silicon is not a carbide forming element, but by lowering the Si content the volume of primary carbides is reduced. Since primary carbides act as crack initiating points, low Si content gives H13M higher toughness and higher strength at elevated temperature, but these two parameters increase at the expense of its machinability. To examine the effect of the increased toughness on machinability, machining tests and cutting force measurements were performed in H13 and H13M (Papers II and IV).

Further, to investigate if the toughness is the cause for the reduced machinability, H13M was heat treated to achieve the same lower toughness as that of H13, and machining tests were performed also on this material (Paper IV).

4.2 Oxidation – Hypothesis 2 (Paper III)

A second hypothesis was proposed where the reduced machinability was explained by the differences in the oxidation properties associated with the changes in the alloying composition. The results of the oxidation experiments support the assumption that the relatively poor machinability of H13M compared to that of H13 is related to a higher oxidation rate and an increased Cr content at the interface between the superficial Fe oxide and the tool steel. Thus, the composition of the tool material surface approaches stainless steel, which is generally known to be difficult to machine [24].
we know from experience, machining stainless steel with high Cr content is difficult, partly because of its high sticking ability to the cutting edge. Machining in H13M could be compared to machining in stainless steel as the composition of the superficial layer of adhered H13M resembles that of stainless steel.

To test this hypothesis, the oxidation properties of three different hot work tool steel including H13 and H13M, and two plastic mould steels (see Table 1) were studied and compared. Together they have different combinations of silicon and chromium content, which is thought to play the most important role in the oxidation characteristics.

4.3 Austenitizing temperature – Hypothesis 3 (Papers IV and V)

One of the reasons for the reduced machinability may be the fact that machining may generate high enough temperatures in the tool steel to initiate phase transformation from ferrite to austenite. Austenitic steels are generally more difficult to machine due to their high sticking ability to the cutting edge [8, 24]. The lower austenite-to-ferrite transformation temperature, A1, of H13M could be one explanation behind its worse machinability. In combination with the fact that austenite has totally different mechanical and thermal properties than ferrite [27, 28], the difference in A1 between H13 and H13M explains the higher cutting forces and higher tendency for built-up edge (BUE) formation.

To test this hypothesis an estimation of the temperature region for austenite formation in the two steels was done by using ThermoCalc [25], comparing with continuous cooling transformation diagram (CCT) [29, 30], and practical Gleeble compression tests at the temperatures around A1 (Papers IV and V).
5 Summary of results

This chapter presents the main results of the five papers in this thesis.

5.1 Machinability of steels with varying silicon content (Paper I)

Silicon is one of several alloying elements that affect the mechanical properties of steels. Carbide precipitation is decreased by reducing the silicon content that results in higher toughness and ductility of the steels. At the same time it was found that machinability of new developed tool steels is reduced. Different types of machining tests were performed in hot forming tool steels with varying Si content, see Paper I.

![Fig. 9: Tool life as a function of Si content when machining with coated CC (Paper I).]

It was shown that the silicon content strongly influences the machinability of the investigated tool steels, see Fig. 9. It seems that the lower silicon content makes the work material (tool steel) more reactive to the CC cutting edge during machining operations and enhances adhesion to the cutting tool. This causes fragments of the cutting edge to be torn out and carried away by the chips. Surface coated CC tools are also affected by this strong bonding and the coating can be partially torn away during an early stage in the machining process. This irregular wear, caused by adhesion must be considered as one of the main features when machining tough and hard tool steels.
5.2 Influence of toughness on machinability - Hypothesis 1 (Papers II and IV)

The first hypothesis tested the observation that the toughness of the steels correlated to the machinability, i.e. increasing the toughness results in a lower machinability. To test this hypothesis the mechanical testing and wear mechanisms were examined in H13 and H13M, see Paper II.

Investigation of toughness and strength revealed significant difference in absorbed energy in Charpy-V tests at elevated temperature and in compressive stresses at 850°C and all strain rates, see Fig. 10.

![Fig. 10: Absorbed impact energy for H13 and H13M at 20-600°C (a) and stress-strain relations in H13 and H13M at 850°C and strain rate 1 s⁻¹ (b).](image)

The fracture toughness test $K_{JC}$ revealed higher $K_J$ value in H13M at temperatures from 200 to 400°C, which means that H13M had higher fracture toughness in this temperature range. At higher temperatures H13 showed a somewhat higher value of fracture toughness than H13M, see Fig. 11a.

![Fig. 11: Fracture toughness $K_{JC}$ (a) and tensile tests (b) of H13 and H13M at elevated temperatures.](image)

Tensile tests at elevated temperature in H13 and H13M revealed a very small difference in 0.2% yield strength and ultimate tensile strength between these two steels in the 20-600°C interval, see Fig. 11b.
Cutting force measurements showed that both the feed force and cutting force were higher in H13M than in H13, see Figs 12a-b. The higher cutting forces were correlated with the observation of early edge fractures for H13M. Larger amounts of H13M also adhered to the cutting tool edge.

Further to investigate if the higher toughness of H13M explains its poor machinability this steel was heat treated to the same lower toughness as that of H13 (see Table 4). Cutting force measurement and wear tests were then performed to compare the machinability between H13M and H13M-L.

Table 4: Charpy-V impact energy for H13, H13M and H13M-L

<table>
<thead>
<tr>
<th>Tool steel</th>
<th>H13</th>
<th>H13M</th>
<th>H13M-L</th>
</tr>
</thead>
<tbody>
<tr>
<td>Test temperature, °C</td>
<td>RT</td>
<td>RT</td>
<td>RT</td>
</tr>
<tr>
<td>Impact energy, J</td>
<td>22</td>
<td>46</td>
<td>28</td>
</tr>
<tr>
<td></td>
<td>500</td>
<td>500</td>
<td>500</td>
</tr>
</tbody>
</table>

As it was expected the cutting forces in H13M-L were reduced to about the same level as that of H13, see Figs 12a-c.

Surprisingly, H13M-L showed an equally poor machinability as ordinary H13M, see Paper IV.

SEM examination of the chips collected from the force measurement tests revealed adhered tool fragments on the exit part of H13M chip. Tungsten was detected by EDS from the entire surface of H13M chips, but not from H13.

From the TEM studies it was clear that the mechanism of wear of the uncoated CC material is the removal of the Co matrix followed by a mechani-
cal plucking of the WC grains. The carbides seem to fracture into smaller pieces before they are incorporated in the flow of tool material, see Fig.13.

![Fig. 13: Representative TEM images obtained in the scanning transmission mode (STEM) of the interface between CC and adhered H13 (a) and H13M (b). The chip flow is from left to right.](image)

There was a significant difference between the two hot work steels as to the structure of the adhered layer, see Fig.14.

![Fig. 14: a) A typical TEM micrograph of the interface between the CC material and adhered H13 steel. The chip flow is from left to right. White areas in the CC material correspond to cavities generated during the CC disruption. b) The characteristic layered structure of adhered H13M.](image)

A study of wear evolution was performed on TiN coated CC when cutting H13 and H13M in interrupted turning using the same cutting parameters for both steels. The cutting was interrupted after 10 s, 1 min, 3 min and 5 min for studies of the edges by 3D white light interferometry.

Generally, a non-uniform adhesion of work material was found along the rake surface. The adhesion and the resulting built-up layer differed between
H13 and H13M. The adhered material was observed in a zone a short distance from the cutting edge on the edges used in H13. After 10 s machining in H13M, adhered material was also found on the rake face. An increase in cutting time led to the formation of a crater, which was filled by adhered material, see Figs. 15.

Fig. 15: 3D white light interferometry of the cutting tool rake face after 10 s (a, b) and 5 min (c, d) machining in H13 and H13M.

5.3 Relation between oxidation properties and machinability – Hypothesis 2 (Paper III)

The second hypothesis related to whether the different oxidations properties associated with the change of alloying composition lie behind the different machinability. The assumption is that a higher oxidation rate and a Cr enrichment in the interface between the superficial Fe oxide and the tool steel play an important role.
To test this hypothesis, the machinability and oxidation properties of three hot work tool steels, H13, H13M and QRO90 Supreme, and two plastic mould tool steels, Stavax ESR and Stavax Supreme, were studied and compared in Paper III. The alloying composition of the tested steels is presented in Table 1.

The stainless plastic mould tool steels were chosen due to their same high Cr content and large difference in Si content. Stavax ESR is easier to machine than Stavax Supreme, see Fig. 16a.

QRO 90 Supreme is relatively easy to machine in spite of its relatively low Si content, see Fig. 16b. This steel has a Cr content of less than half of that of H13 and H13M, and should not be able to form a Cr-rich sub layer under its oxide, which, if proved, would support the hypothesis.

The milling tests showed a lower machinability of QRO90 Supreme than that of H13, but significantly higher than for H13M, see Fig. 16b.

![Fig. 16: Tool life in a) Stavax ESR and Stavax Supreme and b) in H13, QRO90 Supreme, H13M and H13M-L. Note that the cutting parameters differ between a) and b)](image.png)

The samples of the investigated steels were oxidised in air in a conventional furnace at 600°C for 10 min. The surface of the milled H13 and H13M samples and the corresponding chips were also investigated.

The elemental depth profile of the oxidised hot work tool steels characterised by GDOES revealed a superficial iron oxide and an enrichment in elements with high affinity to oxygen, e.g. Cr and Si in the interface to the substrate. The oxide thickness of H13, QRO90 Supreme and H13M was about 0.5, 1.0 and 2.0 µm respectively, see Fig. 17a-c.

Stavax ESR and Stavax Supreme displayed an oxide thickness of about 0.05 µm and 0.07 µm, respectively, also with a strong enrichment of predominantly Cr and Si in the interface to the substrate, see Fig. 17d-e.

The milled H13 and H13M samples revealed an oxide layer of less than 0.01 µm thick, see Fig. 18. Si enrichment was clearly seen below the oxide on the H13 sample.
Fig. 17: GDOES elemental depth profiles of the oxidised surfaces (600°C, 10 min) of H13 (a), H13M (b), QRO90 Supreme (c), Stavax Supreme (d) and Stavax ESR (e).
Fig. 18: GDOES elemental depth profiles of the milled surfaces of H13 (a) and H13M (b).

XPS performed on the surfaces of the H13 and H13M chips revealed similar differences in oxide thickness and chemical elements between these steels as those found by GDOES, see Figs. 17-19. The oxide layer was thicker on the H13M chips. The Si content of H13M is too low to be resolved by XPS.

Fig. 19: Depth profiles of the surfaces of the cutting chips of H13 (a) and H13M (b).

As seen in Fig. 14, H13M steel adhered to the cutting edge exhibited a layered structure with sharp regions of columnar grains and equi-axed grains. A line scan analysis of the adhered steel also revealed that the Cr content was higher and the Fe content was lower in dark areas of the adhered H13M, see Fig. 20.
5.4 Influence of austenitizing temperature – Hypothesis 3 (Papers IV and V)

The third hypothesis is related to the fact that austenite has different mechanical and thermal properties than ferrite. A low A1 for H13M would explain the observed higher cutting forces and higher tendency for BUE formation in this steel.

An investigation including Gleeble testing and ThermoCalc at narrow temperature intervals was performed to fully monitor the ferrite-to-austenite transformation region for the two steels H13 and H13M. Finally, slight changes in the alloying content of H13M (see Table 1) were investigated with ThermoCalc with the aim to find a chemical composition that would increase its A1 and thereby improve its machinability.

No significant differences in compressive flow stress were detected for H13 and H13M in the temperature intervals 600-800°C and 900-1000°C. However the two steels displayed a significant difference in compressive stress in the 800-900°C interval, see Fig 21.
The thermodynamic calculations revealed that there was a difference in austenite content between H13 and H13M in the interval 820-950°C, see Fig. 22.

It proved possible to slightly increase A1 with small changes in Cr, Mo, Mn and Co in H13M, see Fig. 23.

Fig. 21: Examples of Gleeble test curves at 820°C (a) and 870°C (b).

Fig. 22: Calculated austenite content for H13 and H13M estimated by ThermoCalc.

Fig. 23: Austenite content for four hypothetic alloys in Table 1, as estimated by ThermoCalc.
5.5 Thermal properties of H13 and H13M and their influence on the cutting temperature

There are only small differences in thermal properties between H13 and H13M as revealed by Laser Flash, see Fig. 24a-c. The thermal diffusivity was higher at lower temperatures and specific heat was somewhat lower in H13M than in H13. The thermal conductivity $\lambda$ was somewhat higher in H13M up to 600ºC. Then $\lambda$ dropped to lower value compared to H13, see Fig. 24c.

According to the FEM simulations, the highest temperatures were reached on the rake face of the cutting tool. H13M revealed a higher edge temperature, 889ºC compared to 863ºC for H13, see Fig. 25.
Fig. 25: The computed temperature distribution close to the rake face during orthogonal cutting in H13M (a) and H13 (b).

According to the simulation a 10% increase in conductivity lowers the maximum temperature by 24ºC. Increasing the specific heat or the density by 10% had even greater impact on temperature. In both cases the maximum temperature dropped by 43 ºC.
When machining in modified H13 steel (H13M) a significantly worse machinability has been observed compared to traditional H13 steel. The difference in Si content (H13M – 0.06 wt.%; H13 – 1.03 wt.%) has been identified as the most decisive parameter behind this observation.

The aim of this thesis is to explain how Si influences the machinability of these two steels.

6.1 Silicon content

The experiments reported in Paper I and Paper III show that the silicon content strongly influences the machinability of forming tool steels. A reduced Si content affects machinability in a negative way. Several other investigations of the effect of Si on machinability revealed the same result. U. Masahide et al. [31], M. Umino et al. [32], T. Fujii et al. [33] and L-G. Nordh et al. [34], cp. Fig.9, found a dramatic decrease in tool life when machining in H13 with a reduced Si-content.

It seems that the lower silicon content of steels makes the material more reactive to cemented carbide [34]. The cutting edge and the surface coatings are affected by the strong bonding between the tool and the work material and can be partially torn away during an early stage in the machining process. The adhesion results in irregular wear, which is one of the main features when machining in this type of material (Paper I).

6.2 Hypothesis 1 – Toughness

Mechanical testing performed in H13 and H13M, showed that H13M was a much tougher material than H13, cp. Fig. 10a, and that it had a considerably higher resistance against compression at 850ºC, cp. Fig. 10b. The higher toughness and flow stress in compression of H13M should result in higher cutting forces and a higher temperature of the cutting zone (as also revealed by the colour of the chips), cp. Fig. 4 in Paper II. A decreased Si content results in improved ductility of H13 steels due to the reduction in carbon activity, which gives a lower volume of primary carbides, and, consequently, a higher toughness.
The higher cutting forces for H13M were associated with earlier edge fractures. The larger amounts of H13M that adhered to the cutting tool edge, cp. Fig. 8 in Paper II, correlates with the higher edge temperature caused by the higher toughness and flow stress. This excessive adhesion of H13M to the cutting edge is probably associated with a stronger tendency for the BUE formation, which promotes edge fracture and wear. Stronger fluctuations in force for H13M, cp. Fig. 4 in Paper II, especially at tool entrance were probably caused by higher sticking ability of this steel due to its higher toughness and cutting edge temperature.

TEM studies of the wear mechanism of the uncoated cemented carbides revealed the removal of the Co matrix followed by a mechanical plucking of the WC grains. The carbides seem to fracture into smaller pieces before they are incorporated in the flow of the work material, cp Fig.13. There was no principal difference in wear mechanism between the two steels, but the higher edge temperature of H13M gave a higher rate of edge material removal. A significant difference in the structure of the adhered layer between H13 and H13M was revealed. The higher toughness and higher edge temperature of H13M resulted in a layered structure and somewhat larger grains in the BUE, cp. Fig.14.

The conclusion at this stage is that there may be three different reasons behind the lower machinability of H13M.

- Its higher toughness,
- its different reactivity due to lower Si-content,
- its higher resistance to compression at high temperatures

To investigate if the higher toughness of H13M is the main reason behind its poor machinability, this steel was heat treated to the same lower toughness as that of H13 (cp. Table 1) where it is designated H13M-L. Cutting force measurements and tool life estimates were performed. As expected the cutting forces were reduced in H13M-L, cp. Fig. 12, but the machinability still remained poor, cp. Fig. 16b. Consequently, there is no direct relation between toughness and machinability between the investigated steels.

6.3 Hypothesis 2 – Oxidation

Oxidation properties of steels may have a significant influence on their machinability. During machining the work material adheres to the cutting edge. The adhered steel is oxidized between successive engagements of the tool edge. If a steel has a higher oxidation ability, the oxide formed on the adhered layer will be thick and relatively unstable. New layers of work material adhere to the cutting edge in the subsequent cutting engagements, gradually increasing the thickness of the adhered layer. When a critical thickness is reached, the BUE breaks away, and there is a high risk that it simultane-
ously removes fragments of the cutting tool. This results in an accelerated tool wear. If a steel has a lower oxidation ability, then a thin and more stable oxide forms on the adhered layer. This oxide can act as a protection to the cutting tool.

One possibility is that the difference in machinability between H13 and H13M lies in their different chemical composition, which results in a difference in chemical reactivity, primarily in the oxidation properties.

A hypothesis that the different oxidation properties of the steels with different silicon content may be significant for the machinability was tested, cp. Paper III. The poor machinability of H13M may be related to its higher oxidation ability. The addition of Si to the tool steel reduces the oxidation rate. Consequently, the relatively low Si content in H13M gives this steel a higher oxidation ability, which results in the formation of a thicker oxide on the surface. From the GDOES investigation, it is clear that thicker Fe oxides formed on the surfaces of H13M when the samples were oxidised. A significant difference in the oxide thickness and in the alloying composition at the oxide-steel interface on the oxidised H13, H13M and QRO90 Supreme samples was revealed by GDOES, cp. Fig. 17a-c. The properties of the oxides are affected by their alloying composition. The oxide on steel with high Si content will contain silicon because of higher affinity of Si to oxygen compared with Cr. It is also possible that Si in the oxide in the adhered steel acts as a lubricant [35] that reduces the wear of the cutting tool due to the formation of a stable and less thick oxide layer [36, 37] with smearing properties.

The investigations confirmed that machining in H13M could be compared to machining in austenitic stainless steel since the composition of the superficial layer of the adhered H13M resembles that of such steel [38]. Also, the hypothesis about the relation between the oxidation properties of the hot work steels and their machinability can still be valid.

6.4 Hypothesis 3 – Austenite formation

The third hypothesis tested relates to the assumption that machining generates high enough temperatures locally in the work material to initiate phase transformation from ferrite to austenite in H13M but not in H13. This hypothesis is also based on the fact that austenitic steels are usually more difficult to machine than ferritic due to their high sticking ability, high fracture elongation and low thermal conductivity. The ThermoCalc calculations in Papers IV and V revealed a lower start temperature for austenite formation in H13M than in H13, cp. Fig. 22. The thermodynamic calculations showed a good agreement with continuous cooling transformation (CCT) diagram for these steels [29, 30]. The Gleeble tests revealed a deviation in compressive stresses between H13 and H13M in the 820-900°C interval, cp. Fig. 21, which coincides with the temperature interval for ferrite-to-austenite trans-
formation. In this interval the austenite phase (FCC) requires higher compressive stresses than the ferritic phase (BCC) for the same amount of deformation. The reason behind this phenomenon is probably that materials with FCC structure have a tendency to strain-harden at a higher rate than materials with BCC structure [39].

The results of the FEM calculations in § 5.5 also indicate that the temperature at the cutting edge is of the order of the austenitization temperature.

The deformation in Gleeble tests and the deformation during chip formation in machining should be related. Higher compressive stresses lead to higher cutting forces.

Question; is it possible to change the alloying composition of H13M in such a way that its positive tool properties are maintained, but its A1 temperature is increased. The ferrite stabilising elements Cr, Mo and Co and austenite stabilising element Mn were identified as potential candidates for this investigation. ThermoCalc simulations indicated that Mn had the most significant influence on the A1. By reducing Mn content from 0.5 to 0.05 wt. % the A1 temperature could be increased from about 820 to about 845ºC, cp. Fig. 23.

6.5 Thermal properties

Thermal properties of materials may have a considerable effect on their machinability. A large amount of the heat is generated at the tool-work material interface. High cutting temperatures result in thermal softening of cutting tools and promotion of diffusion of chemical elements between work material and tool. The latter is usually named chemical tool wear. High cutting temperatures also increase adherence of work material to a cutting edge. A low thermal conductivity in the steel creates higher temperatures in the cutting region [13, 14]. Machining of steels with low thermal conductivity demands cutting tools with an ability to withstand high temperatures. Materials that conduct heat well facilitate the dissipation of heat away from the area that is being machined. Therefore, high thermal conductivity generally aids the machining process.

Higher thermal conductivity in combination with faster heat transport reduces wear of the cutting tool.

The steel with high thermal diffusivity (D) would have advantage when speaking about machining. Fast transport of the heat from the surface in steels with high thermal diffusivity results in lower cutting temperature.

Studies of the thermal properties of H13 and H13M displayed small differences between these steels, cp. Fig. 24a-c. Thermal properties of the materials are strongly depending on the alloying content [40]. The relatively small differences in thermal properties gave a significant difference in temperature of the tool steels close to the cutting edge, cp. Fig. 25.
increase in the specific heat or density of the tool steel has an even greater impact on temperature, changing the thermal properties by alloying could possibly be an alternative towards improved machinability.

6.6 Comparison between the hypotheses

Three hypotheses concerning the reduced machinability in the low silicon tool steel have been examined. All investigated properties of the tool steel influence the machinability.

Usually, higher toughness and flow stress of a material lead to higher cutting forces and higher temperatures in the cutting zone, resulting in higher wear of the cutting edge. The tests performed in this work revealed that toughness was not the primary reason behind the reduced machinability of the low silicon steel. Therefore the toughness hypothesis is ruled out in this case.

When the composition of a work material surface approaches stainless steel, the material becomes difficult to machine. Any difference in the oxidation rate should also influences machinability. A thick and unstable oxide accelerates wear of the cutting tool. A thin and dense oxide rather acts as a protective layer on the cutting tool. Surface analysis confirmed differences in the oxidation properties between H13 and H13M. The low silicon H13 steel revealed higher oxidation rates giving a composition of the surface material similar to that of austenitic stainless steel. Thus, the second hypothesis is of significance.

The third hypothesis considered the possibilities of a phase transformation from ferrite to austenite to occur in the tool steel during cutting. It is known that austenitic steels are difficult to machine. Tests and simulations revealed lower A1 temperature, higher compressive stresses and higher cutting forces in the H13M steel in the temperature interval for austenite transformation. This leads to higher cutting forces and a reduced machinability of this steel.

The third hypothesis is regarded as the most significant as to the explanation of the difference in machinability between H13 and H13M.

Reducing the Mn content in H13M could be a solution to obtain some improvement in machinability of this grade.
7 Conclusions

In this study the influence of silicon on the machinability of high performance tool steels was investigated by comparing two hot work tool steels. The following conclusion can be drawn.

1. The modified version of H13, H13M, with only 0.06% Si is more difficult to machine compared to the conventional H13, with 1.03% Si.

2. Generally, reduced Si content in H13 type of tool steels results in lower machinability.

3. The low Si content in H13M improves its toughness, reduces the oxidation resistance, and lowers the ferrite-to-austenite transformation temperature. A combination of these properties is the reason for the reduced machinability. Of these, the reduction in austenite temperature, A1 and A3, is probably the most important.

4. Heat treating H13M to the same low toughness level as that of H13 did not improve its machinability, even though the cutting forces were reduced almost down to the level of the latter.

5. The reduced silicon content in H13M results in a thicker oxide and in Cr enrichment in an intermediate layer under the Fe oxide. Increased silicon content in hot work tool steels leads to Si enrichment in an intermediate layer under the Fe oxide, which acts as a lubricant and protects the cutting edge.

6. Gleeble tests show a higher deformation resistance in compression in the 820-900°C-interval in H13M compared to H13.

7. The thermodynamic software ThermoCalc revealed that the transformation from ferrite to austenite occurs at about 50°C higher temperature for H13 compared to H13M.
8. Consequently, the lower temperature transformation from ferrite to austenite in H13M explains the higher cutting forces in this steel.

9. Higher cutting forces give a higher temperature in shear zones of the cut material and of the cutting edge. Increased temperature also promotes adhesion of tool steel to the cutting edge, which is detrimental to the tool wear resistance. In addition, austenite generally has a stronger tendency to adhere to cutting edges than ferrite.

10. In addition, a reduced thermal conductivity in the austenitic state further increases the temperature at the cutting edge.

11. One way to improve the machinability of H13M would be to increase its temperature for transformation from ferrite to austenite without changing its good properties as hot work tool steel.

12. ThermoCalc indicated that addition or subtraction of several alloying elements would change this temperature interval. Of the alloying elements in H13M, the most effective change would occur if the Mn content was reduced. However, changing the composition of any element in H13M would probably also affect other properties.

13. Another approach would be to change the thermal properties by changing the composition.


Den primära orsaken till den försämrade skärbarheten hos H13M tros vara den lägre kiselhalten; 0.06 vikt-% jämfört med 1.03 vikt-% i H13. Framförallt sker en kraftig adhesion mellan spånor och skäverktyg för den lågkiselhaltiga H13 versionen vid bearbetning med hårdmetallverktyg. Skäreggen och eventuella ytbeläggningar på skäverktyget kan brytas loss på grund av kraftiga adhesionseffekter som uppstår vid de höga tryck- och temperaturpå-
känningar som råder vid skärande bearbetning av formstål, speciellt om dessa är i hårdat tillstånd.


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