Anna Medvedeva

Tool steel for tool holder applications

microstructure and mechanical properties
Anna Medvedeva

Tool steel for tool holder applications

microstructure and mechanical properties
Abstract

Large improvements in cutting tool design and technology, including the application of advanced surface engineering treatments on the cemented carbide insert, have been achieved in the last decades to enhance tool performance. However, the problem of improving the tool body material is not adequately studied.

Fatigue is the most common failure mechanism in cutting tool bodies. Rotating tools, tool going in and out of cutting engagement, impose dynamic stresses and require adequate fatigue strength of the tool. Working temperatures of milling cutter bodies in the insert pocket can reach up to 600°C depending on the cutting conditions and material of the workpiece. As a result, steel for this application shall have good hot properties such as high temper resistance and high hot hardness values to avoid plastic deformation in the insert pocket of the cutting tool. Machinability of the steel is also essential, as machining of steel represents a large fraction of the production cost of a milling cutter.

This thesis focus on the improvement of the cutting tool performance by the use of steel grades for tool bodies with optimized combination of fatigue strength, machinability and properties at elevated temperatures.

The first step was to indentify the certain limit of the sulphur addition for improved machinability which is allowable without reducing the fatigue strength of the milling cutter body below an acceptable level. The combined effect of inclusions, surface condition and geometrical stress concentrator on the fatigue life of the tool steel in smooth specimens and in tool components were studied in bending fatigue.

As the fatigue performance of the tools to a large extent depends on the stress relaxation resistance at elevated temperature use, the second step in this research was to investigate the stress relaxation of the commonly used milling cutter body materials and a new steel developed within the project. Compressive residual stresses were induced by shot peening and their response to mechanical and thermal loading as well as the material substructures and their dislocation characteristics were studied using X-ray diffraction.

Softening resistance of two hot work tool steels and a newly developed steel was investigated during high temperature hold times and isothermal fatigue and discussed with respect to their microstructure. Carbide morphology and precipitation as well as dislocation structure were determined using transmission electron microscopy and X-ray line broadening analysis.
Preface

The research associated with this licentiate thesis has been carried out at Uddeholm Tooling AB and the Department of Materials Engineering, Karlstad University. In completing this thesis I am thankful to many people who have supervised and supported me.

First of all, I would like to thank my supervisor Prof. Jens Bergström for his skilful guidance and support. I am deeply grateful to my industrial supervisor Staffan Gunnarsson for his experienced guidance, persistent help and encouragement.

Special thanks to all my colleagues at the Research and Development department at Uddeholm Tooling AB, especially Jörgen Andersson for fruitful discussions and practical support.

My sincerely gratitude also goes to my co-supervisor Pavel Krakhmalev for his valuable advices, to Christer Burman for the technical assistance and to Marianne Johansson for her kind help in administrative issues.

Uddeholm Tooling AB and Sandvik Coromant are gratefully acknowledged for the financial support. Sincere gratitude is extended to Dan Höglund for shot peening of specimens at Sandvik Coromant.

Finally, I wish to thank my family, Ruslan, Natalia and Anatoly for their great support and encouragement.
List of papers

Paper I
A. Medvedeva, J. Bergström, S. Gunnarsson
Inclusions, stress concentrations and surface condition in bending fatigue of H13 tool steel
Steel Research International 5, 2008, 376

Paper II
A. Medvedeva, J. Bergström, S. Gunnarsson, D. Höglund
Stress relaxation resistance for improved fatigue performance of shot peened tool components
Proceedings of the 10th International Conference on Shot Peening, 15-18 September, Tokyo, Japan

Paper III
A. Medvedeva, J. Bergström, S. Gunnarsson, J. Andersson
High temperature properties and microstructural stability of hot work tool steels
To be submitted to Materials Science and Engineering A

Paper IV
A. Medvedeva
Properties of tool steel for tool holder application – A literature review
Karlstad University Studies, 2008
Contents

Abstract 1

Preface 3

List of papers 5

1. Introduction 9

2. Material and experiments 13
   2.1. Hot work tool steels 13
   2.2. Tool steels used in this study 14
      2.2.1. Heat treatment 14
      2.2.2. Microstructure in experimental ingots (Paper I) 15
      2.2.3. Microstructure in production ingots (Paper II and III) 16
   2.3. Experimental 17
      2.3.1. Room temperature bending fatigue (Paper I) 17
      2.3.2. Stress relieving procedure (Paper II) 18
      2.3.3. Hot hardness procedure (Paper III) 19
      2.3.4. Temper resistance procedure (Paper III) 19
      2.3.5. Isothermal high temperature fatigue (Paper III) 19
      2.3.6. Evaluation methods (Paper I, II and III) 19

3. Experimental results and discussions 21
   3.1. Influence of sulphur content on the fatigue strength of tool holders and smooth specimens (Paper I) 21
   3.2. Stress relieving resistance of tool steels for tool holder applications (Paper II) 26
   3.3. High temperature properties and microstructural stability of tool steels for tool holder applications (Paper III)
      3.3.1. Hardness-temperature relationship 30
      3.3.2. Microstructure and hardness at high temperature hold times 31
      3.3.3. Microstructure and softening resistance at isothermal fatigue 32

4. Conclusions 35

Bibliography 37
1. Introduction

There is a growing challenge facing cutting tool manufacturers, tool users and those who provide surface engineering solutions to enhance tool performance. Over the past decades there has been a marked shift in workpiece material utilisation. More easily machinable materials such as cast steel, grey cast iron, tin and copper have been replaced by those with enhanced material properties but which are more difficult to machine, such as tool steel, titanium and nickel based alloys. Components are trending to be machined dry at high speed and leaving very little oversize material. Subsequent the demand on the cutting tools in terms of performance, precision, quality and reliability has been increased.

Large improvements in cutting tool design and technology, including the application of advanced surface engineering treatments on the cemented carbide insert, have been achieved in the last decades. However, the problem of improving the tool body material is not adequately studied.

There are many ways in which a cutting tool body can be damaged, for example, fatigue, chip wear and plastic deformation. But, fatigue is probably the most common failure mechanism in cutting tool bodies, Figure 1. Rotating tools, tool going in and out of cutting engagement, impose dynamic stresses and require adequate fatigue strength of the tool. Fatigue crack starts in the critical radius located in the insert pocket and propagates along the insert pocket resulting in the tool failure.

Different publications [1-3], as well as internal investigation at Uddeholm Tooling AB show that the working temperatures of milling cutter bodies in the insert pocket can reach up to 600°C depending on the cutting conditions and material of the workpiece. There is a trend to machine the materials in pre-hardened condition at high speeds that raises the temperature in the cutting zone. As a result, steel for this application shall have good hot properties such as high temper resistance and high hot hardness values to avoid the plastic deformation in the insert pocket of the cutting tool.

In addition to the properties mentioned above, the machinability of the steel is also essential, as machining of steel represents a large fraction of the production cost of a milling cutter, Figure 2. Its complex shape with flutes and insert pockets as well as small thread holes and small and deep holes for cooling channels require time consuming and advanced machining operations.
Internal investigation of different milling cutters at Uddeholm Tooling AB showed that low-alloyed steels are mostly used in milling cutter bodies. Steel grade SS2541 (Swedish standard) is the most common one for this application. Hot work tool steel AISI H13 is also used by some tool producers but experienced to be difficult to machine. To improve the fatigue strength of the tool body the producers introduce compressive stresses into the insert pocket by different methods. Machining in pre-hardened condition is widely used as a production method, some producers use nitriding or shot peening as a surface treatment of tool bodies.

Figure 1. Fatigue failure on cutting tools.

Summing up, the appropriate steel for the tool body is essential for the functionality and manufacturability of a tool. The detailed analysis of the cutting tool producer’s needs has led to the conclusion that the following properties of the tool steel are the basic factors influencing the usability of the milling cutter bodies: fatigue strength, machinability and high temperature properties.

Figure 2. The component costs in % of total product cost for the typical indexable insert milling cutter produced in the pre-hardened condition. Machining accounts for 63% of the total cost.

Tool steel design problems usually involve conflicting, multiple criteria. Among the different ways of improving machinability, increasing sulphur content in the steel is by far the most common. Sulphur forms manganese sulphides in steels that contribute to easier machining by chip embrittlement, tool protection and flow zone improvements [4]. Unfortunately, the deleterious effect of inclusions on fatigue strength is well known. Adding to this, there is the rough machined surface, often with complicated
geometry and small radii, creating large stress concentrations in the insert pocket of the milling cutter bodies. As the effect of inclusion content on fatigue strength is usually evaluated on smooth polished specimens, the combined effect of inclusions, surface condition and geometrical stress concentrator is not adequately studied. Another example is that it is generally agreed that surface compressive residual stress improves fatigue strength [5-8]. However, residual stresses tend to relieve at higher temperatures. The studies show that there is relatively few literature available concerning the mechanisms of stress relaxation and method of its improving. Furthermore, as the high temperature properties are of vital importance in milling cutter applications, tool steel with pronounced temper resistance and microstructure stability is needed. Secondary hardening steels such as hot work steels are commonly used for tools subjected to thermal exposure [9]. However, these high alloyed steels are generally difficult to machine because they promote a high wear rate of the cutting tool [10].

This thesis mainly aims to improve the cutting tool performance by use of steel grades for the tool bodies with the optimized combination of fatigue strength, machinability and properties at elevated temperatures. The first step was to indentify the certain limit of the sulphur addition which is allowable without reducing the fatigue strength of the milling cutter body below an acceptable level. Effect of inclusions on the fatigue life of the tool steel has been studied in both smooth specimens and in tool bodies containing a milled critical stress concentration. Bending fatigue testing was performed with an experimental AISI H13 tool steel of varying additions of sulphur and oxygen. Manganese sulphide inclusions dominated in the steels containing also oxide inclusions. As the fatigue performance of the tools to a large extent depends on the stress relaxation resistance in elevated temperature use, the second step in this research was to investigate the stress relaxation of the milling cutter body material. SS2541, THG2000 (Uddeholm designation, AISI H13 modified) and a newly developed within the project MCG2006 were used as test materials. Compressive residual stresses were induced by shot peening and their response to mechanical and thermal loading as well as the material substructures and their dislocation characteristics were studied using X-ray diffraction (XRD). The tested steels exhibited different stress relaxation resistance and, consequently, different fatigue strengths during use.

The third paper involves the evaluation of the microstructural and high temperature properties of hot work tool steels THG2000 and QRO90, commonly used for hot work application, and a newly developed tool steel MCG2006. Carbide morphology and precipitation as well as dislocation structure were determined using transmission electron microscopy (TEM) and XRD after tempering, high temperature hold times and isothermal fatigue of the steels. The list of papers is completed with a brief literature review where the essential properties for the tool holder application mentioned above were characterized with the emphasis to the steel type, its microstructure and purity, processing and surface treatment and strong interrelations between them.
2. Material and experiment

2.1. Hot work tool steels

Hot work tool steels are used for hot forging, extrusion and die casting due to their resistance to impact and softening during repeated exposure in hot working operations. Elevated temperature strength and temper resistance are important steel properties for milling cutter applications as well, because milling cutters are subjected to high loads and temperatures during use.

Hot work tool steels, AISI type H, fall into groups that have chromium, tungsten, or molybdenum as the major alloying element. Tool steel heat treatment involves austenitizing, martensite formation, and double tempering. The goal of this processing is to produce a microstructure of tempered martensite with dispersed alloying element carbides. Secondary hardening or the precipitation of alloy carbides at high temperature tempering has the important function of increasing the hardness and retarding the softening. This provides an increased temper resistance so that an appropriate strength can be achieved after tempering at 500-600°C. A good secondary hardening effect is achieved by strong carbide forming elements such as chromium, molybdenum, tungsten and vanadium. The peak hardness associated with secondary hardening increases with increasing alloy content and depends on the balance of primary carbides, retained austenite and the composition of martensite in the as-quenched condition [11].

The type of secondary carbides is sensitive to the specific alloying elements present in the steel. In tool steels containing chromium, the sequence of precipitation with increasing tempering may be $M_7C_3$, then $M_2C_3$, followed by $M_23C_6$. In molybdenum alloyed steels the sequence might be $M_7C_3$, then $M_2C_3$, followed by $M_6C$. Multiple alloying can change the sequence of carbide precipitation. For example, the amount of $M_23C_6$ formed during overaging and the rate of replacement of MC by $M_23C_6$ can be decreased by lower chromium content and increased molybdenum content in H13 steel grade [12].

Other measurements of mechanical performance of hot work tool steels include tensile properties and fatigue strength. To a large extent, the fatigue limit depends on the tensile strength (irrespective of whether strengthening is achieved through cold work,
transformation hardening, solid-solution alloying, or precipitation hardening). Room temperature fatigue strength increases with increase in tensile strength [9, 13]. Fatigue strength is reported to improve from the refinement and volume reduction of primary carbides. Little decrease in strength and no decrease in hardness occur in hot work tool steels at testing temperatures below tempering temperatures. Investigations in isothermal fatigue have shown that the fatigue life is strongly connected to the softening. Retarding the thermally or plastically induced softening would result in a better fatigue life [14].

2.2. Tool steels used in this study

Four different types of tool steels were tested in this study: the most commonly used for tool holder application SS2541, two hot work tool steels with Uddeholm designations THG2000 (AISI category H13 modified for improved machinability) and QRO90 and a newly developed within the project MCG2006 (called HWX in Paper II). Their chemical compositions are listed in Table 1.

Table 1. Chemical composition in wt. %

<table>
<thead>
<tr>
<th>Steel grade</th>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>Cr</th>
<th>Ni</th>
<th>Mo</th>
<th>V</th>
</tr>
</thead>
<tbody>
<tr>
<td>SS2541</td>
<td>0.37</td>
<td>0.3</td>
<td>0.70</td>
<td>1.4</td>
<td>1.40</td>
<td>0.2</td>
<td>0.06</td>
</tr>
<tr>
<td>THG2000</td>
<td>0.39</td>
<td>0.9</td>
<td>0.40</td>
<td>5.3</td>
<td>0.15</td>
<td>1.2</td>
<td>0.90</td>
</tr>
<tr>
<td>QRO90</td>
<td>0.38</td>
<td>0.3</td>
<td>0.75</td>
<td>2.6</td>
<td>0.15</td>
<td>2.3</td>
<td>0.90</td>
</tr>
<tr>
<td>MCG2006</td>
<td>0.30</td>
<td>0.3</td>
<td>1.20</td>
<td>2.3</td>
<td>4.00</td>
<td>0.8</td>
<td>0.80</td>
</tr>
</tbody>
</table>

In Paper I an experimental THG2000 tool steel of varying cleanliness was studied using ingots with intentional addition of sulphur and oxygen. Six different ingots A-F were made with increasing contents of sulphur and oxygen according to Table 2.

Table 2. Sulphur and oxygen contents of test variants of THG2000 (Paper I).

<table>
<thead>
<tr>
<th></th>
<th>A</th>
<th>B</th>
<th>C</th>
<th>D</th>
<th>E</th>
<th>F</th>
</tr>
</thead>
<tbody>
<tr>
<td>S, wt %</td>
<td>0.007</td>
<td>0.023</td>
<td>0.051</td>
<td>0.074</td>
<td>0.090</td>
<td>0.170</td>
</tr>
<tr>
<td>O, ppm</td>
<td>5</td>
<td>28</td>
<td>48</td>
<td>62</td>
<td>56</td>
<td>55</td>
</tr>
</tbody>
</table>

2.2.1. Heat treatment

All steels were tested in hardened and tempered conditions. The hardening treatment was performed in a vacuum furnace and subsequent tempering was made to reach a similar hardness of 44-45 HRC in all tested grades. Hardening and tempering conditions are presented in Table 3.
Table 3.
Heat treatment of the tested materials.

<table>
<thead>
<tr>
<th>Steel grade</th>
<th>Austenitizing</th>
<th>Tempering</th>
</tr>
</thead>
<tbody>
<tr>
<td>THG2000, MCG2006</td>
<td>1020°C/30 min</td>
<td>600°C/2x2h</td>
</tr>
<tr>
<td>QRO90</td>
<td>1030°C/30 min</td>
<td>625°C/2x2h</td>
</tr>
<tr>
<td>SS2541</td>
<td>850°C/60 min</td>
<td>450°C/1h</td>
</tr>
</tbody>
</table>

2.2.2. Microstructure in experimental ingots (Paper I)

In THG2000 experimental ingots the microstructure of the finished samples consists of tempered martensite and depending on steel cleanliness some inclusions, Figure 3. An increasing amount of inclusions was clearly observed as S- and O-content increased, note the difference between the clean and S-rich materials in Figure 3a and b, respectively.

![Figure 3](image1.png)

Figure 3. Microstructures after heat treatment. (a) steel A, (b) steel F with manganese sulphides marked.

Polished plate-samples of test variants A-F of THG2000 were examined in an optical microscope for quantitative evaluation of inclusions using automatic image processing. The area investigated was about 100 mm$^2$ using 200 different fields examined at the magnification of 200x. The inclusion size was characterized by the thickness in the radial direction of the bar. The quantitative evaluation of sulphide inclusions differentiates these all steels in the right order according to their sulphur contents, Figure 4a. Steel A is the cleanest from oxide inclusions following by steel C and the quantitative evaluation of oxides in the other steels is similar, Figure 4b.
2.2.3. Microstructure in production ingots (Paper II and III)

Microstructure of THG2000, MCG2006, QRO90 and SS2541 steel grades in hardened and tempered condition consisted of thin martensite laths with dispersed carbides. The scanning electron microscope (SEM) investigations revealed that the primary carbides in THG2000 and QRO90 were spherical vanadium rich MC carbides, molybdenum rich M₆C carbides and chromium rich M₇C₃ carbides, Figure 5a. MCG2006 contained vanadium rich MC and molybdenum rich M₆C primary carbides, Figure 5b.

Vanadium rich MC, molybdenum rich M₂C and iron rich M₅C secondary carbides, were observed in THG2000 and QRO90 steel grades in hardened and tempered conditions during TEM investigations of the carbon extracted replicas. MC type secondary carbides were found in MCG2006. As expected, Ni is not a carbide-forming element, and, consequently, no modification in the type of precipitates in MCG2006
was noticed. TEM investigations of thin foils revealed that MC type carbides were 2-7 nm, 2-6 nm and 3-4 nm in size for THG2000, QRO90 and MCG2006, respectively. Figure 6. \( \text{M}_2\text{C} \) carbides were found to be 1-3 nm for both steel grades. However, MCG2006 contained fewer of MC carbides than THG2000 and QRO90. A summary of carbides found in the steels is presented in Table 4.

Figure 6. Dark field TEM image showing (a) MC, (b) \( \text{M}_2\text{C} \) type precipitates in THG2000 and (c) MC precipitates in MCG2006 in hardened and tempered conditions.

Table 4.

| Carbides found in THG2000, MCG2006 and QRO90 in hardened and tempered condition 45 HRC. Cross indicates that this type of carbide is present. |
|---|---|---|---|---|---|---|---|
| Primary carbides | Secondary carbides |
| MC | \( \text{M}_4\text{C} \) | \( \text{M}_7\text{C}_3 \) | MC | \( \text{M}_2\text{C} \) | \( \text{M}_3\text{C} \) | \( \text{M}_7\text{C}_3 \), \( \text{M}_{23}\text{C}_6 \) |
| THG2000 | x | x | x | x | x | (2-7 nm) | (1-3 nm) | Not investigated |
| MCG2006 | x | x | x | x | | (3-4 nm, few) | Not investigated |
| QRO90 | x | x | x | x | x | (2-6 nm) | (1-3 nm) | Not investigated |

2.3. Experimental

2.3.1. Room temperature bending fatigue (Paper I)

Fatigue testing was first performed on tool components, Figure 7a, specially designed for fatigue testing. The critical radius of fatigue failure is 1 mm located in the pocket. Subsequently, smooth bending fatigue specimens, Figure 8a, were prepared from the shafts of the fatigue tested tool components. The specimens were designed to fail at the 8 mm transition radius fine ground along the shaft with emery paper (1200 mesh) as final operation. Six test variants of THG2000 with varying sulphur contents were used as test materials.

Bending fatigue tests were performed at room temperature using an Amsler 2 HFP 421 pulsator, Figures 7b and 8b, running with 80 Hz load frequency and 60 MPa preloading. A three dimensional model of the tool component was developed to evaluate the local stresses using the finite element method. A load was applied to the sector area, Figure 7b, and the shaft of the tool component was prescribed rigidly
fastened. Young’s modulus and Poisson’s ratio were 210 GPa and 0.3, respectively. Local stresses in the smooth specimens were calculated using stress concentration factor $k=1.15$ to multiply the nominal stresses [15]. The fatigue test results of the tool components were evaluated according to two methods. The fatigue strength was evaluated using a standard procedure of the stair case method with 50% probability of survival [16] at a fatigue life of $2\cdot10^6$ cycles and 15 test pieces per test variant. For assessment of the lower life range, the fatigue tests were carried out on two stress levels, the minimum number of test pieces was not less than six per stress level. The scatter in life range on one stress level was determined by taking as a basis the normal Gaussian distribution in the Gaussian probability net [16]. The prediction reliability for the probabilities of failure was chosen to be 50%. The evaluated data for the tool components is plotted in Wöhler diagrams. As to smooth specimens, only the fatigue strength at $2\cdot10^6$ load cycles was evaluated using the stair case method and 10 specimens per test variant.

Figure 7. (a) The tested tool components. Length - 110 mm, diameter of the top - 19 mm, (b) Loading of tool components, circular mark shows the crack initiation site, the critical radius in the pocket is 1 mm.

Figure 8. The smooth bending fatigue specimen. (a) geometry, (b) loading.

2.3.2. Stress relieving procedure (Paper II)

Stress relief heat treatments were performed at different temperatures (200-700°C for 2 hours) on shot peened flat samples of SS2541, THG2000 and MCG2006. Shot peening was made according to the following conditions: pressure 4 bar, shot direction angle $90^\circ$, nozzle distance from surface $75\pm5$ mm, peening intensity $15\pm2$A (Almen A) with a 100% coverage. Shot peening media used was super conditioned cut wire with diameter of 0.35 mm and hardness of 700 HV. Residual stress analysis including
determination of surface stress and depth distribution was accomplished following
treatments at the different temperatures.

2.3.3. Hot hardness procedure (Paper III)

The hot hardness is determined by measuring the indentation produced by pressing a
square-based diamond pyramid against the specimen surface at high temperature. The
Cohen High-Temperature Micro-Hardness Tester, adopting the micro-Vickers method
with the standard E92–82 (1997), consists of a high temperature microscope, a weight
loading unit and a heating chamber connected to a vacuum pump, which permits
heating without heavy oxidation. The operating load for hardness testing was 300
grams.

Cylindrical test specimens ∅11 x 30 mm of THG2000, QRO90 and MCG2006 were
hardened and tempered according to the procedures described in 2.2.1. During hot
hardness testing they were heated up to elevated temperature and held for 15 min.
Three indentations were made on the surface, and the steel hardness was determined
from the size of the mean value indentation. Then the temperature was increased and
the procedure was repeated. The temperature levels used were 100, 200, 300, 350, 400,
450, 500, 550, 600, 650 and 700°C.

2.3.4. Temper resistance procedure (Paper III)

Small samples 10 x 10 x 30 mm of THG2000, QRO90 and MCG2006, hardened and
tempered according to the procedure described above in 2.2.1., were tempered for
different times 1, 2, 5, 10, 20, 30, 40, 50 and 100 hours at different temperatures of
500, 550, 600, 650°C. Hardness measurements followed the tempering procedure.

2.3.5. Isothermal high temperature fatigue (Paper III)

Isothermal high temperature fatigue (ITF) tests were carried out in air using a 100 kN
servo hydraulic INSTRON testing machine in push-pull type load mode. Symmetrical
tests were performed in strain control (+0.5%/-0.5%) at both 450 and 550°C. The tests
had a sinusoidal strain wave shape, 0.5 Hz cycle frequency and tensile start direction.
Testing time was about 2 hours for all steel grades. Cylindrical test specimens with 6.5
and 20 mm waist diameter and length, respectively, were used. The specimens were
hardened and tempered according to the procedure, described above in 2.2.1, and
turned in hard condition. The fatigue tests were run to failure at approximately 1500-
3000 load cycles, corresponding to approximately 1-2 hours test time. Particular care
was taken to evaluate the cyclic softening throughout the testing to determine the
resistance to stress relaxation.

2.3.6. Evaluation methods (Paper I, II and III)

SEM equipped with energy dispersive X-ray spectrometer (EDS) was used to examine
the specimens and tool fractured surfaces as well as the chemical analyses of
inclusions (Paper I). The microstructure of the steels as well as primary carbide
morphology was also analysed (Paper III).

By using X-ray diffraction (XRD) it is possible to determine the amount of
macrostresses and microstrains. The macrostresses provides information on the
residual surface stress, so that the stress relief resistance of the steels was examined (Paper II). Residual stresses were evaluated by XRD using Cr-Kα radiation on the (211) martensite/ferrite planes in a Seifert XRD 3000 PTS X-ray diffractometer, operating at 40 kV and 35 mA.

The microstrain arises from the distortion in the microstructure, created mainly by dislocations, but also by carbides and alloying elements. XRD line broadening analysis for microstructural state was accomplished by an integral breadth method. Separation of the size and strain contributions to the line broadening was made by deconvolution of their intensity distributions (simple and squared Lorentzian, respectively). Thus, the coherently diffracting domain size and the average root mean square strain variation within the grain, the microstrain, and their variation during high temperature loading were determined (Paper II and III). The dislocation density expressed as a proportional value [17], the ratio between microstrain, \(<\varepsilon^2>^{1/2}\), and domain size, D, was estimated from:

$$\rho = \frac{2 \sqrt{3}}{b} \frac{<\varepsilon^2>^{1/2}}{D}$$

where \(b\) is the Burger’s vector.

TEM was used to study the size and type of secondary carbides, which are of vital importance to evaluate the microstructural stability of steels (Paper III). The TEM examination was performed on carbon extraction replicas and on thin foils by bright and dark field imaging. The replicas and foils were prepared according to [18].
3. Experimental results and discussions

3.1. Influence of sulphur content on the fatigue strength of tool holders and smooth specimens (Paper I)

The aim of this study is to investigate the combined effect of inclusions, surface condition and geometrical stress concentrator on the fatigue life of the tool steel in both smooth specimens and in tool components containing a milled critical stress concentrator.

Large improvements in fatigue life of tool steels have been achieved in the last decades by reduction of the inclusion content [19]. Among a number of factors affecting the harmfulness of different inclusion types, inclusion size transverse to loading direction is probably the most important parameter [19-24]. The composition of the inclusion is also of significance, globular calcium alumimates appear to be more detrimental than elongated manganese sulphides [21]. Inclusion shape is also of importance since it influences the stress field inside and around the inclusion. Smooth inclusions elongated in direction of the major principal stress cause less stress concentration in the surrounding matrix than sharp edged inclusions [19].

Since fatigue failure may initiate from the component surface, surface irregularities such as machining marks pose a great threat to fatigue crack nucleation. Smoother surfaces give improved fatigue life, while fatigue life of rough surfaces deteriorates [24-26].

Residual stress pattern induced by machining is critical for component fatigue life, it is generally agreed that surface tensile stress reduces fatigue strength and compressive residual stress improves it [5-8].

Many components contain stress raising notches, grooves or other geometrical stress concentrators, which increase nominal stress from cyclic loading with a stress concentration, quantified by the stress concentration factor $k_t$. The geometry of the critical area is an essential factor controlling the component fatigue behaviour in the presence of the stress concentrator which detrimental effect on fatigue life of critical components is the subject of much intensive work.
In the present work bending fatigue testing was performed with THG2000 (an experimental AISI H13 tool steel modified) of varying additions of sulphur and oxygen.

Longer fatigue life and higher fatigue strength are displayed for the steel A with lowest sulphur and oxygen contents, Figure 9. It is significantly better than other steels (compare for instance to steel F). In the range of 0.02 – 0.09 wt% in sulphur content, the fatigue lives of the tool components are rather similar, as well as the fatigue strength of the steels, Figures 9 and 10.

In case of smooth specimens, the difference in fatigue strength is much larger than in tool components, Figure 10. The fatigue strengths for steels E and F were not estimated for lack of specimens, but, as all specimens were broken at the stress level of 1000 MPa, supposed to be lower than the fatigue strength of the steel D.

Some fractography study using SEM determined the cause of failure and the size, type and location of inclusion in case it initiated failure. On all fracture surfaces it was possible to identify the location of crack initiation. Typical morphology of fatigue fracture surfaces for the steels with lowest and highest sulphur contents is presented in Figure 11.

Seven crack initiating inclusion conglomerations and six surface defects due to machining could be identified out of the 90 fractures of the tool components. The initiating inclusions were in all cases Al-, Si-oxides situated at the specimen surface, no sulphide inclusions were detected. Inclusion conglomeration areas were larger than 20 µm in size. The remaining 77 fractures initiated at the surface but without any obvious defects.

When testing smooth specimens, crack initiating inclusions were found on the majority of the fracture surfaces, whereas for the tool holders fractures initiated mostly at the surface. In those cases where inclusions were identified as crack initiators they were situated very close to the specimen surface. Putting together size of the failure initiating inclusions and the applied stress for failure, Figure 12, the trend is towards decreasing size for higher failure stress. Here, the oxide type inclusions dominated as crack initiators and the effect of sulphides was smaller. Moreover, the sulphide type inclusions initiated cracks at higher stress level than oxides, Figure 12.

![Figure 9. Wöhler diagrams for the tool components made from steels with different sulphur content (A – 0.007%S, B – 0.023%S, C – 0.051%S, D – 0.074%S, E – 0.090%S, F – 0.17%S). Each data point is an average estimate of six failures.](image-url)
Figure 10. Comparison between the fatigue strength obtained for smooth specimens and the tool components made from steels with different sulphur content (A – 0,007%S, B – 0,023%S, C – 0,051%S, D – 0,074%S, E – 0,090%S, F – 0,17%S). Brackets show the standard deviations.

Figure 11. Typical fracture morphology of fatigue fracture surfaces. (a) and (b) smooth specimens, steel A – 0,007%S and F – 0,17%S respectively, (c) and (d) tool components, steel A and F respectively. Initiation is indicated with arrows.
In testing of both the tool components and the smooth specimens, surface initiation of failures is natural due to the bending load mode. Fatigue cracks for the smooth specimens appear in the plane perpendicular to the long direction and propagate in radial direction. Contrary in the tool components the failure crack runs in the transverse plane. The latter is more detrimental as the inclusions are elongated in the axial direction, which is the rolling direction. As well, the radius in the pocket of the tool component is a strong stress concentrator enhancing the crack initiation localisation. The small radius in the tool component results in a high stress concentration and a small critical stressed volume. Consequently, the probability of an occurring critical failure defect is smaller. Hence, it explains the low fraction of inclusion initiations in the tool components contrary to the smooth specimens where we have a larger critical stressed volume and failure controlled by inclusions. Surface stress raisers are also of importance in the tool components since its milled surface is rough, characterized by $R_{\text{max}}=15\mu\text{m}$, which is comparable to some inclusion sizes of the steels.

Residual stresses are another important factor to consider when improving the fatigue strength. The crack initiation site of parts requires an adequate level of compressive residual stress within the machined surface to increase the fatigue strength. In the present case a beneficial contribution of -80 MPa and -130 MPa residual stresses measured with XRD were present in the smooth specimens and the tool components, respectively.

Thus, when there is low demand on the surface roughness and in the presence of a critical stress concentrator, the effect of sulphur content on fatigue strength is lower. Generally, large oxide inclusions should be avoided, thus demanding low oxygen content in the tool steel. Also, sulphur reduces impact toughness [22], hence it should be within controlled limits.

The findings of the present study suggest that in a real component there is a balance between these different factors. In particular it implies that the addition of sulphur up to a certain limit, here about 0.1%, for improved machinability, is allowable without reducing the fatigue strength below an acceptable level.

End milling and drilling machinability tests were carried out on the test variants A-F of H13 material, showing the machinability improvement with increasing sulphur.
content, Figures 13 and 14. Lower flank wear of the milling cutter, longer tool life and greater number of drilled holes is experienced when the steel becomes rich in sulphur.

![Graph showing flank wear vs. milled length in end milling of A-F tests variants of H13 material. Each point is the average between three tests. Material hardness is 45 HRC.](image1)

Figure 13. Flank wear vs. milled length in end milling of A-F tests variants of H13 material. Each point is the average between three tests. Material hardness is 45 HRC.

![Graph showing number of drilled holes in A-F test variants of H13 material. Each value is the average between three tests. Material hardness is 45 HRC.](image2)

Figure 14. Number of drilled holes in A-F test variants of H13 material. Each value is the average between three tests. Material hardness is 45 HRC.
3.2. Stress relieving resistance of tool steels for tool holder applications (Paper II)

Earlier investigations on the factors influencing the fatigue life of the indexable insert milling cutters [27] showed that compressive residual stresses in the insert pocket of the milling cutter is the most important parameter. Type of material does not have much influence on the fatigue strength if the tool is machined in soft condition and then hardened and tempered leaving no residual stresses in the material. However, the type of material does influence if the tool is machined in pre-hardened condition or shot peened. The more compressive stresses can be introduced in the material, the higher the fatigue strength of the tool.

At higher working temperatures of milling cutters, e.g. up to 600°C, the influence and behaviour of the compressive residual stresses are more uncertain, since they tend to relieve. As the fatigue performance of tools to a large extent depends on the stress relaxation resistance in elevated temperature use, knowledge of stress and microstructure stability is essential for evaluating the service life of a tool. The overall aim of the present study is to optimise the stress relaxation resistance of tool components to achieve better fatigue strength at elevated working temperatures.

Shot peening is an extensively used process in the production of milling cutters to improve their fatigue strength. The response to shot peening induced residual stresses of SS2541 and THG2000 and MCG2006 steels was evaluated with respect to stress relief heat treatments. Not only the residual stresses, but also the material substructure and its dislocation characteristics are of importance. Dislocation structures were determined using X-ray diffraction to explain the preference of the different steel grades.

The shot peening process induces a work hardened surface layer with large compressive residual stresses well below 50µm depth, Figure 15. Initial compressive stress level is equal (-700 MPa) to all three steels, while they during subsequent thermal treatment show different ability to retain the residual stress and work hardening, Figure 15 and 16. Obviously, MCG2006 is the best and SS2541 is the worst in this respect. Particular about THG2000 is its plateau range at 450 to 550°C where stress relaxation seems to stabilize.

By using X-ray line broadening analysis it is possible to measure the amount of disturbance in the matrix. The disturbance primarily comes from dislocations, but also contributed by alloying elements and precipitates, and is expressed as microstrain and domain size of the matrix. A high dislocation density is produced during shot peening of the steels. The substructure and amount of dislocations are believed to be important contribution to the strength and should be maintained during high temperature use. Decrease of microstrain and domain size increase measured by XRD originates from the dislocation density decrease.

In general, dislocation density decreases with temperature and surface depth due to the rapid microstrain decrease and domain size increase, Figure 17a. It was also noted during XRD measurements that initial dislocation density after shot peening is highest in MCG2006 and lowest in SS2541, and the 450 to 550°C plateau range was revealed in terms of dislocation density as well as in THG2000, Figure 17b. This may be related
to remnants of secondary hardening effects occurring during tempering in THG2000. Precipitation of secondary carbides in THG2000 may affect the dislocation reconfiguration (dislocations themselves promote the heterogeneous nucleation of precipitates), increasing stress relaxation resistance in this steel.

There are two primary processes, annihilation and rearrangement of dislocations into lower energy configurations, which explain the stress relaxation at higher temperatures [28]. Different dislocation mobility at higher temperatures in the investigated steels may be explained by different microstructure stability of the steels. During the annihilation and rearrangement of dislocations, the particles in THG2000 and MCG2006 may pin dislocations and thus inhibit their movement, increasing the stress relaxation resistance of these steels comparing with the low-alloyed SS2541. Smaller, finely coherent particles are more preferable for stronger pinning. Moreover, interaction between dislocations and alloying elements and the martensitic phase also influence the stress relaxation and dislocation mobility. The stronger stress relaxation at temperatures above 550°C in THG2000 than in MCG2006 is proposed to be due to more rapid over-aging of carbides.

![Compressive stress profile versus stress relief temperature in tested materials.](image)

Figure 15. Compressive stress profile versus stress relief temperature in tested materials.
Figure 16. Percentage of compressive stresses left on the surface of THG2000, SS2541 and MCG2006 after stress relieving at different temperatures for 2 hours. Initial stresses after shot peening are set to 100%.

Figure 17. (a) Depth distribution of calculated proportional dislocation density for shot peened THG2000 after stress relief heat treatment at 300 and 500°C, (b) proportional dislocation density in THG2000, MCG2006 and SS2541 after stress relieving heat treatment at different temperatures.
3.3. High temperature properties and microstructural stability of tool steels for tool holder applications (Paper III)

Indexable insert tools used in machining operations are exposed to high temperatures and cyclic mechanical loads during use. Such working conditions impose high requirements on the elevated temperature properties of indexable insert material such as hot hardness, temper resistance and high temperature fatigue to avoid deformation in the insert pocket.

Secondary hardening steels such as hot work steels are commonly used for tools subjected to thermal exposure [9]. However, these steels, highly alloyed with strong carbide forming elements as Cr, V, Mo, etc., are generally difficult to machine and machining of such steels represents a large fraction of the production cost of a tool.

Figure 18. Flank wear vs. milling time in end milling of THG2000, MCG2006 and SS2541. Each point is the average between three tests. Material hardness is 33 HRC.

Figure 19. Flank wear after drilling 1000 holes in THG2000, MCG2006 and SS2541. Each point is the average between three tests. Material hardness is 33 HRC.
End milling and drilling machinability tests carried out on THG2000, MCG2006 and SS2541 showed the machinability improvement with decreasing alloying element content in the steels, Figures 18 and 19. Earlier investigations have shown that tool steels experience softening during tempering and high temperature fatigue and that the softening is strongly connected to the microstructure and its stability at high temperature [12, 28-31].

The aim of this work is to evaluate microstructural and high temperature properties of THG2000 and QRO90 tool steels, commonly used for hot work applications, and a newly developed tool steel MCG2006 with leaner balance of carbide alloying elements and added Ni content to improve combined machinability and high temperature strength. Hot hardness and isothermal fatigue (ITF) “short” time tests were carried out conforming the “initial stage/ few engagements” of milling. Relatively long test times in temper resistance experiments correspond to the full life time of a milling cutter body.

3.3.1. Hardness-temperature relationship

In moderate temperature range, up to 0.5Tm, where Tm is the melting point of a metal or alloy, the mechanism controlling the softening is reported to be the dislocation slip and deformation twinning. Resistance of the solute impurity atoms to the moving dislocations gives the main contribution to the flow stress and determines the softening resistance.

All tested materials exhibited softening with temperature increase, Figure 20, where the softening behaviour depends on the steel grade. Up to the temperature of 350°C all steel grades exhibited the continuous hardness decrease with increasing temperature and no difference in softening behaviour was observed in the steels. However, the change in softening mechanism is observed for THG2000 at 450-550°C. The hot hardness curve has a plateau in the temperature range of 450-550°C, and then the abrupt hardness decrease occurred, Figure 20. The hot hardness stabilisation at 450-550°C may be related to remnants of secondary hardening effects occurring during tempering in THG2000.

![Figure 20. Hardness of the steels at elevated temperatures.](image-url)
MCG2006 and QRO90 demonstrated the highest hot hardness values at temperatures more than 600°C, where the softening behaviour is similar for both steel grades. Comparing the hot hardness and ITF tests, described later in 3.3.3., both conducted in a shorter time span, yield similar softening results testing the different steels.

3.3.2. Microstructure and hardness at high temperature hold times

Typical microstructure of hot work tool steels consists of tempered martensite with high dislocation density and alloyed carbides precipitates. The softening resistance of hot work tool steels is determined by the changes in character of alloy-element carbides at elevated temperature and by the recovery of the dislocation-rich martensite structure. At high working temperatures the steels exhibited progressive hardness decrease that is associated with the fine carbide coarsening and dislocation density reduction. These two phenomena are strongly related to tempering time and temperature. Amount of alloying elements in the supersaturated martensitic phase is also of importance as they influence the precipitation kinetic and dislocation rearrangement.

No changes of carbide morphology and size was observed in the test materials after two hour tempering at the temperature of 550°C that is below the tempering temperature of the steels. However, a sharp decrease of hardness takes place during the initial stage of tempering at temperatures exceeding the tempering temperatures of the steels, Figure 21. The carbide coarsening is observed in THG2000 after tempering for 2 hours at 650°C during TEM investigations of the carbon extracted replicas. Then, this short period is followed by a quasi-linear decrease of hardness, which depends on tempering temperature and the steel grade, Figure 21.

Temper resistance test demonstrated a better temper resistance in QRO90 in comparison to THG2000 and MCG2006 at all tempering temperatures, Figure 21. In QRO90 molybdenum content increase promotes M₂C formation and diminution of chromium content delays the transformation from MC and M₂C carbide towards M₂₃C₆ form [12, 32]. In comparison with THG2000, MCG2006 demonstrated a temper resistance of the same level as THG2000 at the tempering temperatures up to 550°C and an improved temper resistance at higher temperatures and longer tempering times. It may be related to the more stable MC carbides because of the lower chromium content in MCG2006 than in THG2000.

Amount of alloying elements in the martensitic phase may be also of influence. Tempering can be considered as a phase transformation promoted by diffusion from an unstable state (martensite) towards a quasi equilibrium state (ferrite + globular carbides) [33]. MCG2006 is highly alloyed with Ni that is known to retard the phase transformation happening in the steel.
Combining thermal effects and mechanical loading in the isothermal fatigue (ITF) test, Figure 22, cyclic straining processes lead to microstructural changes causing cyclic hardening and/or softening of the steel. Softening behaviour was observed in all steels at 550°C and in QRO90 and MCG2006 at 450°C. THG2000 showed hardening behaviour at 450°C.

Researchers have proposed various mechanisms to explain the softening behaviour of steels: resolution of precipitates after being cut by dislocations to a size smaller than the critical size for particle nucleation [34]; over-aging of precipitates [30]; and rearrangement of dislocation substructure into a dislocation subgrain structure of lower internal stress [29-30]. Here, cyclic softening during isothermal fatigue is considered to be caused by the rearrangement of the initial high dislocation density to a lower dislocation density. Microstructural state evaluated by line broadening analysis of X-ray diffraction was investigated on the ITF specimens before testing them in hardened and tempered condition, after isothermal fatigue testing at 450°C and 550°C and after the tempering at 550°C for 2 hours. The major contributions to XRD line broadening, lattice microstrain and domain size, were measured. It was found out that a microstrain decrease and domain size increase occurred after the isothermal tests is similar for all steel grades and depends just on the testing temperature. As a result, the degree of dislocation rearrangement is dependent also only on the testing temperature, Figure 23. Small fine carbides are reported to obstruct the dislocation motion and retard softening [28, 31]. Here, no difference in softening behaviour of MCG2006 was observed in

![Figure 21. Hardness dependence on time for the steels tempered at different temperatures.](image-url)
comparison with THG2000 and QRO90, although it contained fewer precipitates than
the two others, Figure 6. However, Ni in the supersaturated martensitic phase in
MCG2006 can also be considered to influence the dislocation rearrangement. No
coarsening of carbides was observed in the tested steels after ITF at 550°C. It may be
because the testing time was not enough for dissolution of smaller carbides and growth
of the bigger ones.
The cyclic hardening in THG2000 at 450°C may be the result of the remnant
secondary carbide precipitation as the hardness of the steel is increased after the
fatigue testing. The similar effect was observed in THG2000 during stress relaxation
and hot hardness testing in the same temperature range, Figures 16, 18 and 20.
Comparing the proportional dislocation density in the steels after ITF and two hour
tempering at 550°C, Figure 23, it was found that the dislocation density decreased a lot
more after ITF testing than it would have if the specimens were only subjected to
thermal exposure. This indicates a mechanical softening in the steels.

![Figure 22. Normalized stress amplitude $\sigma_a/\sigma_{a,0}$ versus cycle number for THG2000, MCG2006 and QRO90 in ITF performed in strain control $\pm0.5\%$ at 450 and 550°C. Each test series is averaged from three specimens. Final failure occurred for all steels and temperatures at 1500-3000 loading cycles.](image-url)
Figure 23. Proportional dislocation density of THG2000, QRO90 and MCG2006, steels in hardened and tempered condition, after ITF testing at 450°C and 550°C and after 2 hour tempering at 550°C. Brackets show the standard deviations.
4. Conclusions

The following conclusions can be drawn from this thesis:

1. There is a potential to improve the tool holder performance combining all the demanding properties such as improved fatigue strength, high temperature properties and machinability and optimizing them by better understanding the effecting parameters and mechanisms involved.

2. In tool holder application where demand on surface roughness is low and geometrical stress concentration is present the addition of sulphur to the steel up to a certain limit, for improved machinability, is allowable without reducing the fatigue strength below an acceptable level.

3. Different tool steels exhibit a variety of stress relaxation resistance that depends on their microstructure, temper resistance, and the working temperature. The decrease of dislocation density is one of the mechanisms involved in the stress relieving of different steels. Hot work tool steels showed to be more preferable before the low alloyed tool steels in tool holder application because of their ability to inhibit the rearrangement and annihilation of dislocations induced and, consequently, higher fatigue strengths during use. A modified hot work tool steel with decreased Cr content and added Ni content improved fatigue life due to enhanced stress relaxation resistance.

4. The initial stage of softening of different hot work tool steels is controlled by the dislocation rearrangement and annihilation and depends on the working temperature. The different modifications in steel composition presently tested have no influence on the dislocation density decrease. The long time softening is affected by the temper resistance of the tool steel that strongly depends on the carbide morphology and their over-ageing resistance. A modified hot work tool steel leaner alloyed with carbide forming elements and added Ni has better temper resistance than standard H13 steel grade at higher temperatures and longer tempering times. Greater Mo content and lower Cr content than that in standard H13 promoted the best resistance to softening among tested grades at all temperatures.
Bibliography

Large improvements in cutting tool design and technology, including the application of advanced surface engineering treatments on the cemented carbide insert, have been achieved in the last decades to enhance tool performance. However, the problem of improving the tool body material is not adequately studied.

Fatigue is the most common failure mechanism in cutting tool bodies. Rotating tools, tool going in and out of cutting engagement, impose dynamic stresses and require adequate fatigue strength of the tool. Working temperatures of milling cutter bodies in the insert pocket can reach up to 600°C depending on the cutting conditions and material of the workpiece. As a result, steel for this application shall have good hot properties such as high temper resistance and high hot hardness values to avoid plastic deformation in the insert pocket of the cutting tool. Machinability of the steel is also essential, as machining of steel represents a large fraction of the production cost of a milling cutter.

This thesis focus on the improvement of the cutting tool performance by the use of steel grades for tool bodies with optimized combination of fatigue strength, machinability and properties at elevated temperatures.

The first step was to indentify the certain limit of the sulphur addition for improved machinability which is allowable without reducing the fatigue strength of the milling cutter body below an acceptable level. The combined effect of inclusions, surface condition and geometrical stress concentrator on the fatigue life of the tool steel in smooth specimens and in tool components were studied in bending fatigue.

As the fatigue performance of the tools to a large extent depends on the stress relaxation resistance at elevated temperature use, the second step in this research was to investigate the stress relaxation of the commonly used milling cutter body materials and a new steel developed within the project. Compressive residual stresses were induced by shot peening and their response to mechanical and thermal loading as well as the material substructures and their dislocation characteristics were studied using X-ray diffraction.

Softening resistance of two hot work tool steels and a newly developed steel was investigated during high temperature hold times and isothermal fatigue and discussed with respect to their microstructure. Carbide morphology and precipitation as well as dislocation structure were determined using transmission electron microscopy and X-ray line broadening analysis.