Chromium martensitic hot-work tool steels

Chromium martensitic hot-work tool steel (AISI H13) was developed to endure the severe conditions of high temperature metal forming operations such as die casting, hot rolling, extrusion and hot forging. The mechanical properties are high and strongly connected to the microstructure and have been improved over the years by alloying and heat treatment. Damages still occur and one of the most common failure mechanisms is thermal fatigue.

In this thesis the thermal fatigue damage on hot forming tools has been studied. Several types of hot work tools steels have been experimentally tested and the microstructural changes during thermal fatigue have been evaluated. The tool material behaviour has also been simulated to support the integration of die design, tool steel properties and use.

The general aim of this thesis is to increase the knowledge of the chromium martensitic hot-work tool steel damage, performance and microstructure.
Johnny Sjöström

Chromium martensitic hot-work tool steels
– damage, performance and microstructure

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PREFACE

The work presented in this doctoral thesis has been carried out at the Department of Materials Engineering, Karlstad University, and at the Uddeholm Tooling AB research department. The financial support Böhler Uddeholm AG is gratefully acknowledged.

First of all I would like to express my gratitude to my supervisor Prof. Jens Bergström for all the guidance and support. Secondly I would like to extend my thankfulness to the research department and research group at Uddeholm Tooling AB for all the support, especially Jörgen Andersson. Also, I would like to thank my colleagues at the University of Karlstad who has supported me with fruit full discussions and practical guidance. I would also like to thank a small group at Luleå University for the co-work in simulation and modelling.

Finally, I would like to thank my wife Malin and my children William and Stina for supporting me and giving motivation.

*It is easier to do the work than to explain why you have not done it.*

Martin van Buren
List of enclosed papers

This doctoral thesis comprises the following papers, referred to in the text by their roman numerals:

**Paper I**  J. Sjöström and J. Bergström  

**Paper II**  J. Sjöström and J. Bergström  

**Paper III**  J. Sjöström and J. Bergström  

**Paper IV**  J. Sjöström and J. Bergström  

**Paper V**  J. Sjöström and J. Bergström  
“Cyclic behaviour modelling of hot-work tool steels at high temperature fatigue”, Submitted to the International Journal of Fatigue.

**Paper VI**  David Hjerts´en, Johnny Sjöström, Jens Bergström and Mats Nässtöm  
“Finite Element Simulation of the Tool Steel Stress Response As Used in a Hot Forging”, 8th International Conference on numerical Methods in Industrial Forming Processes, Columbus, 2004
Other publications

This work has also resulted in a Licentiate thesis and a publication, which is not included in the Doctorial thesis. The Licentiate thesis and that publication are listed below.


- Johnny Sjöström and Jens Bergström, Thermal fatigue testing of chromium martensitic hot work tool steel after different austenitizing treatments, International Conference on Advanced Material Processing Technology, July 2003 Dublin. Accepted for publication in Journal of Advanced Material Processing Technology.
The author’s contribution to the papers

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

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

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

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

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

Paper IV  Part of planning, writing and evaluation all experimental work.
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1 INTRODUCTION

Hot forming processes are among the oldest and most important metal forming technologies and accounts for a large percentage of fabricated metal products. However, the metal industry today is very competitive and a metal former must carefully evaluate the costs of the operations necessary for converting each material into finished products. Therefore the industry continuously strives to lower the production costs of each operation. The die plays the most essential part in all types of hot forming operations, because it usually gives the object its final complex shape. Since the die usually is expensive to manufacture it has a major influence on the production costs of the products. Some metal workers even claims that a high-quality die with a long lifetime is the key for a successful and cost-effective production. Forming techniques such as hot forging and die casting are two popular ways of forming net and near net shaped components, since they are economical and high-speed methods. Also, modern statistical and computer based process design and simulations are becoming more important than ever in the struggle of reducing the production costs. Simply, because computer based process simulations can optimise the production, without using expensive tooling and testing.

There are many ways in which a hot forming die can be damaged; for example wear, plastic deformation, gross cracking, thermal fatigue and mechanical fatigue [1-3]. But, thermal fatigue (often called heat checking) is probably the most common failure mechanism in all hot forming techniques and may be defined as fatigue produced by the repetition of stresses that are thermal in origin, i.e. stresses that arise because the expansion or contraction from the heating or cooling is constrained. As these stresses accumulate by each repetition they will eventually cause either excessive distortion or thermal fatigue cracking. Thermal fatigue is usually recognised as a network of surface cracks and is commonly facilitated by creep and environmental effects, such as oxidation [4-5]. The thermal fatigue
mechanism can roughly be divided into three stages [6] substructural and microstructural changes, crack nucleation and crack growth.

This thesis mainly aims at improving the die material and its performance, and that problem was encountered in several steps, Fig. 1. The first step was to identify the life-limiting factors of a hot forming die and, since, the conditions of a die varies a lot depending on the application, two different hot forming dies from two different hot forming techniques, hot forging and die casting, was investigated. Several damage mechanisms were found on the dies, such as; wear, oxidation and plastic deformation, but common for both techniques and dies were that thermal fatigue was found to be the most detrimental damage mechanism.

![Work scheme of the different parts of this thesis.](image)

In order to improve a hot-work tool steels thermal fatigue resistance it is essential to understand which properties are important and, nevertheless, which property is the most important. This can be done by testing different types of steels, each with different properties, or to use different heat treatments for the same material. In this study, the aim was not only to determine, the most vital
property, but also to understand why and how. Therefore, the connection between the mechanical properties and the microstructure was studied, using three different types of hot-work tool steels. Two kinds were heat-treated in four different ways, attaining different properties and different microstructures in specimens of the same material. The tool steels were tested in a thermal fatigue test rig, where it was not only possible to determine the thermal fatigue resistance by measuring the crack depth, but also to measure the surface strains with a laser speckle technique. To further investigate the tool materials behaviour during high temperature loading isothermal fatigue testing was performed, mainly to record the fatigue lives and the die materials stress/strain softening behaviour. In some isothermal fatigue tests the same strain condition as measured on the surface of the thermal fatigue specimen was used and in some a life length criteria from a real hot forming application. Several conclusions could be drawn from the experimental results, where, as an example, it was found that in some hot forming conditions the softening and the fatigue life was closely connected to the temper resistance. It was also concluded that the softening could be divided into three parts: a rapid initial softening, a second stable softening and rapid final fracture softening. It also showed that there was a great difference in softening behaviour between the different heat treatment conditions.

The third step in this research was to investigate what the difference in softening and fatigue life originates from. In the four heat treatment conditions different austenitizing temperatures were used creating different microstructures, thus different properties. The heat treatment conditions were characterised by measuring the mechanical properties by fracture toughness tests, tensile tests and temper resistance tests. The microstructure of some materials conditions used in the thermal fatigue testing was also investigated by scanning electron microscopy (SEM), X-ray diffraction and transmission electron microscopy (TEM), to evaluate the role played by microstructural features such as dislocations, carbides and grain size. The SEM was mostly used to examine the specimen’s fractured surface of the different materials and heat treatment conditions. X-ray line broadening analysis was used to measure the microstrains, which mainly arises from the lattice disturbance by dislocations (some disturbance also comes from carbides and alloying), and how it changes during high temperature loading for the different heat treatment conditions. Transmission electron microscopy was used to study the
dislocation structure, and also to determine the amount and type of carbides in the different conditions.

The fourth step is to use the material related data in numerical simulation. Mainly, because hot forming involves several of different temperature and load conditions and with numerical simulation it may be possible to simulate the life and performance of certain technique without using expensive and time consuming tooling and testing. In this work, a non-linear kinematic and isotropic hardening model was used to simulate a hot-work tool material behaviour during specific hot forging and die casting conditions. There are two papers on material modelling; one, which has focused on a specific hot forging operation and where the material behaviour is simulated. The other paper focuses more on the model itself, where the material related parameters are investigated in regard to the microstructure.

This thesis can also aid in the knowledge and development of other important hot forming factors in order to resist tool failure:

1) Design, tool material behaviour during use, involving selection of dimensions, corner radii and section changes.
2) Heat treatment and mechanical properties at different heat treatments, since improper heat treatment is one of the most common causes of failure [1].
3) Dominant damage mechanisms and conditions in a tool.
4) Selection of properties for martensitic chromium hot-work tool steel in order to resist thermal fatigue.
5) Martensitic chromium hot-work tool steel behaviour during high temperature loading and the relation to the microstructure.
6) Kinematic and isotropic hardening/softening parameters relation to microstructure.
2 HOT FORMING

Hot metal forming consists of a forming process either by plastic deformation or solidification where the metal is shaped by tools or dies. The hot deformation process occurs above the metals recrystallisation temperature, which usually is between 0.4 and 0.5 of the materials absolute melting point. At that temperature the metal is easy to shape, since it behaves in a perfectly plastic manner. The metals become neither internally stressed nor work hardened, and an unlimited amount of hot-working can be performed without component fracture.

Metal forming by plastic deformation is probably the oldest forming method. The earliest records of metalworking describe the simple hammering of gold and copper in various regions of the Middle East around 8000 B.C. [7]. In the late Copper age (around 4000 B.C.) it was discovered that hammering of metal brought a desirable increase in strength [8], and a type of hammer forging by hand became a popular why to form the metals. Most of the metal forming was done by hand until the 13th century, when the tilt hammer driven by waterpower was developed. It was mainly used for forging bars and plates. However, rolling miles was not invented until later. It is documented that Leonardo da Vinci, who is also believed to be the first to use this method, rolled flat sheets of precious metals on a hand-operated two roll mill for coin making in 1495 [7]. But, it was not until 200 years later that large mills capable of hot rolling ferrous metals were developed.

During the Industrial Revolution at the end of the 18th century an, almost exponential, increase of hot forming industry occurred. The demands of larger hot formed quantities increased, which resulted in the invention of the high speed steam hammer, with a hydraulic press. Even tough several of new types of forming operations were developed the fundamental technique still remains. Hot forging, Fig. 2(a), is still one the most common hot forming techniques, because it is cost effective, and also, since it gives the final product exceptional mechanical and thermal properties. More than 2 million tons of steel parts are produced each year in Europe by hot forging [9].
Die casting is another forming technique, but instead of forming by plastic deformation, as in the case of hot-forging, the die-casting product is shaped by solidification of melted metal, Fig. 2(b). Exactly when the casting of metals began is not known, but it is believed to have started somewhere between 5000 and 3000 B.C. [2]. However, die casting is a fairly new technique and is characterised by a source of hydraulic energy that pass on high velocity metal into a cold die chamber where the metal is solidified into desired shape. This is a rapid event, with a filling time in the order of seconds [10]. Because of this high velocity filling, die casting can produce objects with complex shapes and thin walls at high rate. The production is of the order of 100 objects per hour depending on the size of the machine.

Fig. 2. a) Hot forging and b) die casting.
3 HOT-WORK TOOL STEELS

Common for the two techniques, hot forging and die casting, is that they both have a die or a tool, which gives the product its final shape. These tools are usually very complex and expensive and in order to lower the production costs, they need to last for a long time. The materials used in the dies for hot forming are today completely made of a special type of steel, called tool steels. The development of tool steel is closely related to the evolution of steels in general, but the beginning of tool steel history is generally regarded as 1740 [11], when Benjamin Huntsman, a clock maker melted pieces of blister steel in a crucible. By melting the steel instead of heating iron in charcoal (made carbon diffuse in into the iron producing blister steel), it made the steel much more homogenous and, thus, stronger. However, modern tool steels, with complex alloying and heat treatments, are much more advanced. But, the understanding of the interrelationships among carbon content, alloy composition and processing, that developed the modern tool steels, came only gradually in the 19th century. The earliest recorded benchmark for the development of modern tool steels is when Robert Mushet in 1868 [11] intentionally added tungsten to high carbon steel. Much has happened in the development since then and today there exist numerous types of tool steels, but the desire to increase the performance of the tool steels still remains.

The steels used for hot forming is a special type of tool steel, made to withstand a combination of heat, pressure and abrasion and has been classified hot-work tool steel, AISI type H. All hot-work tool steels are used in a quenched and tempered condition. The most essential properties for these types of steels are high levels of hot strength, ductility, toughness, thermal conductivity, creep strength, temper resistance and also low thermal expansion [2, 3]. Steels that need to maintain its properties at high temperatures, e.g. hot-work tool steels, require having an increased temper resistance so that an appropriate strength can be achieved after tempering at 550 /650 °C. The most convenient method is to use a secondary hardening reaction involving the precipitation of alloy carbides [2, 3, 12]. A good secondary hardening effect is achieved by strong carbide forming elements.
such as chromium, molybdenum, tungsten and vanadium. These elements play an important role when the tools steel is subjected to high temperatures, since they precipitate as fine alloy carbides, which not only retards the softening but also increases the hardness.

The AISI type H steel is divided into three subgroups named after the dominant alloying element [3]:

**Chromium** hot-work steels (types H10 to H19) are well adapted to hot-work of all kinds. Especially dies for the extrusion of aluminium and magnesium, but also as die-casting dies, forging dies and hot shears.

**Tungsten** hot-work steels (types H21 to H26) are used to make mandrels and extrusion dies for high temperature applications, such as the extrusion of brass, nickel alloys and steel. They are also suitable for use in hot-forging dies of rugged design.

**Molybdenum** hot-work steels (types H42 and H43) are almost similar to tungsten hot-work steel with almost identical characteristics and uses, but have their principal advantage in their lower initial cost. These alloys, especially molybdenum and the low carbon content, make the steel more resistant to heat checking.

### 3.1 HOT-WORK TOOL STEELS INVESTIGATED IN THIS STUDY.

The most commonly used hot-work tool material is AISI category H13, which presently bests fulfils the demanding properties. Three different types of H13 steels where tested in this study, Premium H13 and two with the Uddeholm designations QRO 90 Supreme and DIEVAR. Their respective chemical composition along with the chemical range for the AISI H13 is listed in Table 1. QRO 90 and DIEVAR were heat-treated in four different ways, using four different austenitizing temperatures followed by a tempering to an approximately equal hardness, Table 2.
### Table 1. Chemical compositions in wt. %

<table>
<thead>
<tr>
<th>Steel grade</th>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>Cr</th>
<th>Mo</th>
<th>V</th>
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<tbody>
<tr>
<td>QRO 90 Supreme</td>
<td>0.38</td>
<td>0.30</td>
<td>0.75</td>
<td>2.6</td>
<td>2.25</td>
<td>0.82</td>
</tr>
<tr>
<td>Premium H13</td>
<td>0.39</td>
<td>1.0</td>
<td>0.4</td>
<td>5.2</td>
<td>1.4</td>
<td>0.9</td>
</tr>
<tr>
<td>DIEVAR</td>
<td>0.37</td>
<td>0.20</td>
<td>0.5</td>
<td>5.0</td>
<td>2.36</td>
<td>0.55</td>
</tr>
<tr>
<td>AISI H13</td>
<td>0.32-0.4</td>
<td>0.80-1.20</td>
<td>0.20-0.50</td>
<td>4.75-5.50</td>
<td>1.10-1.75</td>
<td>0.80-1.20</td>
</tr>
</tbody>
</table>

### Table 2. Heat treatment and hardness (all hardness measurements are within ±10 HV)

<table>
<thead>
<tr>
<th>Austenitizing</th>
<th>Tempering 1</th>
<th>Hardness 1</th>
<th>Tempering 2</th>
<th>Hardness 2</th>
</tr>
</thead>
<tbody>
<tr>
<td>QRO 90</td>
<td>625°C/2*2h</td>
<td>480 HV</td>
<td>625°C/2*2h</td>
<td>470 HV</td>
</tr>
<tr>
<td>1020°C/30min</td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>1060°C/30min</td>
<td>625°C/2*2h</td>
<td>560 HV</td>
<td>640°C/2*2h</td>
<td>470 HV</td>
</tr>
<tr>
<td>1100°C/30min</td>
<td>625°C/2*2h</td>
<td>540 HV</td>
<td>640°C/4*2h</td>
<td>470 HV</td>
</tr>
<tr>
<td>1150°C/30min</td>
<td>625°C/2*2h</td>
<td>560 HV</td>
<td>650°C/2*2h</td>
<td>470 HV</td>
</tr>
</tbody>
</table>

<table>
<thead>
<tr>
<th>Austenitizing</th>
<th>Tempering 1</th>
<th>Hardness 1</th>
<th>Tempering 2</th>
<th>Hardness 2</th>
</tr>
</thead>
<tbody>
<tr>
<td>DIEVAR</td>
<td>600°C/2*2h</td>
<td>480 HV</td>
<td>600°C/2*2h</td>
<td>470 HV</td>
</tr>
<tr>
<td>1020°C/30min</td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>1060°C/30min</td>
<td>600°C/2*2h</td>
<td>510 HV</td>
<td>600°C/3*2h</td>
<td>470 HV</td>
</tr>
<tr>
<td>1100°C/30min</td>
<td>600°C/2*2h</td>
<td>530 HV</td>
<td>600°C/3*2h</td>
<td>470 HV</td>
</tr>
<tr>
<td>1150°C/30min</td>
<td>600°C/2*2h</td>
<td>510 HV</td>
<td>600°C/4*2h</td>
<td>470 HV</td>
</tr>
</tbody>
</table>

### 3.1.1 Heat treatment and microstructure

The most stable carbide in this type of steel is the VC-carbide, and because of its stability it plays an important role in the heat treatment. The VC-carbide is responsible for pinning the grain boundary in order to obstruct the grains from growing. Thus, a higher austenitizing temperature can be used without a significant grain growth. A typical example of this can be seen if the average grain size versus austenitizing temperature for QRO 90 and DIEVAR, Fig. 3(a), is compared. QRO 90 contains vanadium closer to stoichiometry and therefore has better resistance to grain growth. In QRO 90 an actual grain growth does not occur until a temperature of 1100 °C is reached, while in DIEVAR it starts before 1060 °C. However, when the temperature in QRO 90 is high enough to dissolve the VC-carbides the material experiences a rapid grain growth. Investigations of the carbides (Paper IV) after different heat treatments showed that DIEVAR contained several VC-carbides after austenitizing treatment at 1020 °C, Fig. 4(a), but no VC-carbides at all were found after austenitizing at 1100 °C.
If a higher austenitizing temperature is used in the heat treatment more of the primary carbides will be dissolved and thus increase the secondary hardening effect. It has been shown that V:C ratios close to stoichiometric gives a better secondary hardening effect as well, simply because the amount of VC available for precipitation is greater [12]. Consequently, the temper resistance is increased. The temper resistance is also dependent on another important feature such as the stability of the carbides. Studies have shown that decreasing the chromium content and increasing the molybdenum will generate more stable carbides, because the chromium rich carbides M₇C₃ and M₂₃C₆ can easily coalesce and coarsen [12, 13].

Tempering resistance test of DIEVAR and QRO 90 (Paper II) at four different heat treatments, 1020, 1060, 1100 and 1150 °C, not only showed that an increased
austenitizing temperature improves the temper resistance, but also that QRO 90, with a lower amount of chromium, had better tempering resistance than DIEVAR, Fig. 3(b). Investigations of DIEVAR showed that the secondary carbides were mainly chromium carbides of the types M-C₃ and M₂₃C₆, Fig. 4(b). After hardening all test material were tempered to an equal hardness 470±10 HV₃₀.

3.1.2 Mechanical properties

In general, the increased austenitizing temperature in the heat treatment improves the temper resistance, but it has detrimental effect on the impact toughness and the ductility, Fig. 5. From the Charpy-V impact test and the tensile test it was found that DIEVAR had the highest impact toughness and that QRO 90 had lower ductility than prem. H13, Fig. 5(d).

![Charpy-V impact toughness versus test temperature at different conditions for (a) DIEVAR, (b) QRO 90 and (c) Premium H13. d) Reduction of area vs. test temperature for Premium H13 and QRO 90.](image)

To obtain the required strength all hot-work tool steels contains 0.3-0.4 wt% carbon, since the strength increases with the amount of carbon [2]. Studies has shown (Paper III) that there is hardly any difference in strength between the different heat treatment conditions, Table 3
<table>
<thead>
<tr>
<th>Test temp.</th>
<th>Yield strength [MPa]</th>
<th>Tensile Strength [MPa]</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>RT</td>
<td>200°C</td>
</tr>
<tr>
<td><strong>QRO 90</strong></td>
<td></td>
<td></td>
</tr>
<tr>
<td>TA 1020</td>
<td>1310</td>
<td>1165</td>
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<td>TA 1150</td>
<td>1445</td>
<td>1320</td>
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<tr>
<td><strong>Prem. H13</strong></td>
<td></td>
<td></td>
</tr>
<tr>
<td>TA 1020</td>
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<td>1125</td>
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<td>TA 1100</td>
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<tr>
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<td>1255</td>
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<tr>
<td><strong>DIEVAR</strong></td>
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<td>1385</td>
<td>1260</td>
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</tr>
<tr>
<td>TA 1150</td>
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</table>
4 EXPERIMENTAL METHOD AND EQUIPMENT

The objective of the experimental study was to simulate the tool materials behaviour during operation. Since it was found in paper I that thermal fatigue was the most detrimental damage mechanism the hot-work tool material were tested and evaluated in a thermal fatigue testing rig. Thermal fatigue was obtained by rapidly changing the temperature on the surface of the test material using induction heating (Paper III). The same temperature intervals and, thus, the same surface strains as found in various operations of hot forming was used. However, the hot-work tool steel was also tested and evaluated in another high temperature fatigue test, i.e. isothermal fatigue, using constant high temperature during fatigue loading, (Paper II, V and VI). Isothermal fatigue testing is used as complement to thermal fatigue and is also needed for material model parameter identification used in the numerical identification.

4.1 THERMAL FATIGUE (PAPER III)

The thermal fatigue testing was performed by rapid induction heating, using 3 MHz high frequency and 25 kW power to induce fast heating close to the surface of the specimen. The cooling was mainly done by internally circulating silicon oil at the temperature of 60 °C through a 3 mm axial hole of the cylindrical specimens with 10 mm diameter and 80 mm length. An external cooling effect is added by an argon gas flow, which as well provides an inert atmosphere. The surface temperature is monitored by a pyrometer, but also measured by a thermocouple. The temperature cycle consists of a steep ramp to maximum temperature followed by a slower cooling to a minimum temperature, Fig. 6(a) and (b) for maximum temperatures 600 and 700 °C, respectively. The maximum and minimum cycle temperatures were chosen so as to simulate different die casting processes [3], Table 4.
Table 4. Thermal fatigue test conditions

<table>
<thead>
<tr>
<th>Cycle designation</th>
<th>Max. temp.</th>
<th>Min. temp.</th>
<th>Time to max. temp.</th>
<th>Total cycle time</th>
</tr>
</thead>
<tbody>
<tr>
<td>Tmax 600</td>
<td>600°C</td>
<td>200°C</td>
<td>0.2s</td>
<td>11.2s</td>
</tr>
<tr>
<td>Tmax 700</td>
<td>700°C</td>
<td>200°C</td>
<td>0.3s</td>
<td>14.3s</td>
</tr>
<tr>
<td>Tmax 850</td>
<td>850°C</td>
<td>200°C</td>
<td>2.2s</td>
<td>26.2s</td>
</tr>
</tbody>
</table>

Since, the surface of an operating tool is only subjected to a limited amount of oxide at maximum temperatures an external argon gas flow was used to provide an inert atmosphere around the specimen.

The surface strains on the specimens were also measured using a non-contact laser-speckle technique. A technique, which is based on an interference pattern in the reflected beam created from the surface topography, detected by four CCD sensors, Fig. 7 and 8. When the surface expands an equal change occurs in the interference pattern and a computer acquires the movements of the interference pattern and is then translated, by a Matlab™ program, to surface strains. More information on the test rig can be found in [10].
4.2 ISOThERMAl FATIGUE (PAPER III, V AND VI)

The other important experimental technique used was isothermal fatigue testing, which is necessary in order to identify a hot-work tool material high temperature fatigue behaviour. In this technique the strains or stress (depending on control mode) is regulated by a computer, which makes it possible to measure the
stress or strain softening during the test. The test is performed on a time glass shaped specimen placed in a 100 kN INSTRON servohydraulic testing machine, Fig. 9. The specimen is subjected to a push-pull type load mode. The strains in the specimen were measured using an extensometer. Tests were carried out in either symmetrical or non-symmetrical loading using either sinusoidal or triangular wave shapes. Heating of the specimen was achieved with a resistive furnace with test temperatures from 200 up to 600 °C, continuously measured with a thermocouple. The isothermal fatigue testing was carried out in air.

4.3 EVALUATION TECHNIQUES (PAPER III, V AND VI)

To learn more about the microstructural changes in the die material, the experimental specimens were evaluated using X-ray diffraction (XRD), scanning electron microscope (SEM) and transmission electron microscope (TEM). By using XRD it is possible to determine the amount of macrostresses and microstrains. The macrostresses provides information on the residual surface
stress, but the main objective was to examine the microstrains in the material. The microstrains arise from the distortion in the microstructure, created by mainly dislocations, but also by alloying elements. By measuring the amount of dislocations at various load cycles it is possible to learn how the microstructure changes. Another important instrument used to evaluate the changes in the material is the TEM, which makes it possible to actually see the dislocations and its structure. It is also possible to see and examine the carbides, which is vital information in order to evaluate the microstructure in the different materials and conditions. Since various heat treatments were used in this study it is important to learn how their individual differences, in microstructure, affects the thermal fatigue resistance. The SEM was mainly used to examine the specimens and tool fractured surface, and since it was equipped with energy dispersive X-ray spectrometer (EDS) it was also used in chemical analyses. The fractured surfaces of specimens from Charpy-V impact tests, tensile test and fatigue test were analysed.
5 CONDITIONS IN THE HOT FORGING AND DIE-CASTING TOOLS (PAPER I)

Since the rapid thermal variations on the tool surface are responsible for the thermal fatigue cracking it is important that the temperature condition is estimated. In order to calculate the strains induced from the thermal cycling and also to set the limits for experimental testing.

5.1 TEMPERATURE CONDITIONS IN HOT FORGING OF STEELS CRANKSHAFT

The tool used in hot forging of heavy duty vehicle Perkins crankshafts was investigated. Prior to forging, the work-piece is preheated to a temperature of 1250 °C and the tool to a temperature of 200 °C. The work-piece is forged with 39 MN of force. As a result of the intimate contact between the hot-work-piece and the tool during the deformation process, heat is transferred very rapidly into the die and produces high surface temperatures. When the work-piece is removed the surface will cool rapidly, and even more when subjected to lubricant spraying. The tool surface temperature is significantly determined by the contact time, which is 0.15 s excluding the post-forge dwell time and the pre-forge time. The surface of the tool was nitrided prior to use, mainly for improved wear resistance. The tool was repeatedly taken out of use to be reconditioned, i.e. ground and renitrided.

The die surface temperature was measured during die opening in the hot forging sequence using an infrared (IR) camera. The measurements revealed that the in the hot forging process demonstrated large temperature variation on the tool surface, Fig. 10. The highest temperatures were at the die cavity edge where there is a long contact time. These hot spots had a maximum temperature of 600 °C approximately 1 second after forging and a minimum of 200 °C after cooling.
Fig. 10. Thermography of the surface of the hot forging tool showing the (a) minimum temperature profile after cooling and (b) maximum temperature profile after forging.

The lowest temperatures were found in the bottom of the crank shaft shape, where it, during one forging cycle, fluctuates between 250 °C, directly after die opening, and 90 °C, after cooling.

5.2 CONDITIONS IN ALUMINIUM DIE CASTING

Two aluminium die-casting dies in a double cavity tool used in the production of flywheels were investigated. One of the dies was prepared with a duplex coating, nitriding + PVD-CrN, prior to use, while the other remained with the machined metal surface. Both dies were inspected after 4000 shots, to determine if the coating had any effect on the crack initiation, but only the non-coated tool was inspected after failure. The casting process was run with an aluminium melt at 670 °C, a 0.02 seconds filling time and a 70 seconds total cycle time. After die opening and ejection of the flywheel, the tool is sprayed with water containing 0.4 % oil cooling the surface ensuing a temperature drop to 220 °C.
6 DAMAGES IN HOT FORMING TOOLS (PAPER I)

Tools from two hot forming processes, aluminium die casting of flywheels and hot forging of steel-crankshaft, were investigated. Common for both these processes are the rapid temperature variations, which gives rise to thermal strains. As these strains accumulate by each repetition they will eventually cause either excessive distortion or thermal fatigue cracking. From the investigation thermal fatigue cracking was found on the surface of the hot forming tools.

6.1 DAMAGES IN CRANKSHAFT HOT FORGING TOOL

Several damages were found on the tool, such as thermal fatigue, wear, plastic deformation and fatigue, Fig. 11. It was also found that the tool failed due to a large fatigue crack, and that the initiation of the fatigue crack, was caused by thermal fatigue cracking on the die cavity edge, Fig. 12. Typical for the thermal fatigue crack location is that the contact time is long and thus the temperature high. A sharp corner in the forging geometry acts as stress raiser and further facilitates the propagation of the crack.

Fig. 11. Damages on a hot forging tool showing (a) thermal fatigue cracking and (b) wear.
6.2 DAMAGES IN AN ALUMINIUM DIE-CASTING TOOL FOR FLYWHEELS

Only two types of damages were found on the die, thermal fatigue and erosive wear, Fig. 14(a) and (b), respectively. However, the dominant damage was thermal fatigue cracking, which was found at almost all sharp corners. After 350,000 production cycles, including service, the tool was taken out of production, since the thermal fatigue cracks had grown too large, causing severe marks in the product.
6.3 THERMAL FATIGUE CRACK INITIATION (PAPER I)

The thermal fatigue crack initiation occurred very early in both the die-casting and hot forging tool. Visual observation of the tool surfaces revealed that thermal fatigue cracking occurred in the hot forging tool after less than 100 production cycles and on the die-casting tool before 4000 shots. Roughly, it means that crack initiation occurred at less than 1% of the tool's life and the remaining is a matter of resistance to crack propagation.

6.4 THERMAL FATIGUE CRACK PROPAGATION (PAPER I)

The propagation of the hot forging crack is most likely driven by both the mechanical forging loads and thermal loads in the hot surface regions, but as the crack propagates away from the hot spots, the driving force for crack propagation will mainly be mechanical. In general, it is believed that the crack propagation was not facilitated by any environmental effect. Even though oxides were observed in the thermal fatigue cracks, the amount was so small that oxide-induced wedge cracking is assumed to be of minor importance, Fig. 16(b). However, in the die-casting case the crack propagation has encountered a different problem, which is liquid metal that fills the crack and then solidifies and enhances the propagation. The crack opening by the wedge mechanism is clearly indicated in Fig. 15(a) and Fig. 16(a). Residues from the lubricant was found in both tools, silicon and carbon, Fig. 16(c) and (d), respectively, but is not believed to aid in the propagation.
Fig. 15. Thermal fatigue crack in the (a) die casting and (b) hot forging tool surface (SEM picture)

Fig. 16. EDS-analysis of a thermal fatigue crack in the die-casting tool surface showing (a) aluminium and (b) oxide and in the hot forging tool surface (c) silicon and (d) carbon is seen.
7 THERMAL FATIGUE TESTING RESULTS
(PAPER I AND II)

From the investigations in paper I the maximum temperature variation was measured during hot forging and since the aluminium melt injected into the die-casting die has a temperature of 670 °C the maximum temperature on the tools surface was roughly estimated to be somewhere between 600 and 700 °C. Thermal fatigue testing was performed using the investigated temperature conditions and, also, the information from previous work of die-casting conditions [10]. The main objective was to evaluate the surface strain condition in the tool [paper I], and also the different heat treatment conditions resistance to crack initiation and crack growth when exposed to rapid cycling to high temperatures [paper II].

Surface strain measurements of the tool steels used in the investigated dies, which was premium H13, showed that total strains caused by thermal cycling [paper I] were approximately $\varepsilon_{\text{tot}} = 0.14\%$ at $T_{\text{max}}$ 600 °C and $\varepsilon_{\text{tot}} = 0.18\%$ at $T_{\text{max}}$ 700 °C, Fig. 17(a). The actual thermal strain is in reality much larger, Fig. 17(b), but since it is constrained by the cooler bulk material it only reaches the amount measured as total strain. However, the constraining caused by the cooler bulk material induces a stress and this stress can be expressed as hypothetical strain, calculated by subtracting the thermal strain from the total strain [paper I], Fig. 17(b). The calculated mechanical strains were found to be approximately $\varepsilon_{\text{mech}} = -0.47\%$ at $T_{\text{max}}$ 600 °C and $\varepsilon_{\text{mech}} = -0.6\%$ at $T_{\text{max}}$ 700 °C. Similar strain results were found for DIEVAR [paper II].
The other objective of the thermal fatigue testing was also to evaluate the influence of the microstructure when exposed to thermal fatigue. Therefore DIEVAR specimens heat treated in four different ways, using austenitizing temperatures ($T_A$) 1020, 1060, 1100 and 1150 °C, were experimentally tested at three different maximum temperatures. The crack depth was used as a measure of the thermal fatigue resistance and it was found that the heat treatment $T_A$ 1100 °C experienced the best resistance to thermal fatigue [paper II], Fig. 18. It was also found that DIEVAR had better thermal fatigue resistance than Premium H13, Fig. 19.
7.1 HARDNESS AFTER THERMAL FATIGUE (PAPER II)

Microhardness tests of the fatigued specimens showed that if the maximum temperature in the thermal fatigue test exceeded 600 °C the surface of the specimens experienced a significant decrease in surface hardness, Fig. 20. It was also found that the heat treatment that experienced the least decrease in hardness also had the best thermal fatigue resistance, Table 5. Therefore it was concluded by looking at the properties for the different heat treatments that as soon as the test temperature comes close to, or above, the materials tempering temperature, the property of tempering resistance becomes increasingly important.

Fig. 20. Hardness profiles of DIEVAR and Premium H13 specimen with the maximum cycle temperature (T_{max}) of (a) 600, 700 (b) and 850 °C tested after 5000 and 10000 thermal cycles.
Table 5. Hardness loss measured 0.01 mm from the surface of DIEVAR austenitized at 1020, 1060, 1100 and 1150 °C and Premium H13 specimen, at $T_{\text{max}}$ 700 °C tested after 10000 thermal cycles

<table>
<thead>
<tr>
<th>Cycle Number</th>
<th>TA 1020</th>
<th>TA 1060</th>
<th>TA 1100</th>
<th>TA 1150</th>
</tr>
</thead>
<tbody>
<tr>
<td>Prem. H13 10000</td>
<td>180 HV</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>DIEVAR 10000</td>
<td>110 HV</td>
<td>90 HV</td>
<td>60 HV</td>
<td>80 HV</td>
</tr>
</tbody>
</table>
During most hot forming techniques the tool properties will decrease with increasing amount of production cycles. The decrease depends on several things both extrinsic e.g. load and temperature and intrinsic e.g. material and microstructural condition. The main objective of this study was to identify the materials behaviour during high temperature loading and to evaluate the microstructural conditions. Three types of materials were tested QRO 90, DIEVAR and Premium H13. Four different hardening treatments with different austenitizing temperatures were selected (1020, 1060, 1100 and 1150 °C), to obtain a set of test materials with different microstructural conditions. However, only one condition of Premium H13, austenitized at 1020 °C, was tested. The strain amplitudes used in the tests of QRO 90 and DIEVAR were chosen as to simulate the mechanical surface strain condition in a die-casting tool [paper I and II]. But, the strains used on the Premium H13 specimens were set to fail within the range of a crankshaft hot forging tool life length i.e. somewhere between 2000 and 10000 load cycles.

All tested materials in this study exhibited a cyclic strain or stress softening, Fig 21, where the softening behaviour showed to be dependent on the temperature, mean strain, strain amplitude, stress or strain control mode, steel grade and heat treatment. The heat treatment proved to have a considerable influence on the softening, where TA 1060 and 1100 °C generally showed a better softening resistance than all other conditions. The results also showed that the resistance to isothermal mechanical softening at 600 °C is higher for QRO 90 than for DIEVAR. However, at 500 °C, both DIEVAR and QRO 90 specimens (treated at TA 1100 °C) indicated softening stabilisation, and only a small difference in the softening behaviour was found, Fig 21(c). The fatigue life of the two materials increased with the austenitizing temperature until TA 1100 and 1150 °C, for QRO 90 and DIEVAR respectively, was used, Fig. 21(d).
The softening can be separated into three stages: a primary, a secondary and a final stage. The primary stage was roughly the same for all test results, but the secondary stage was different and significantly influenced by the heat treatment. An increased austenitizing temperature from 1020 to 1060 or 1100 °C results in decreased secondary softening and in general an increased fatigue life when tested at a temperature of 600 °C.

From the Premium H13 tests it was found that the strain range needed to reach fatigue failure in the specimens within the 2-10000 load cycles were around 1.2 % at temperatures 200-400 °C, 1.0 % at temperature 500 °C and 0.9 % at temperature 600 °C, Fig. 22. It was also found that the primary stage softening was small.
Fig. 22. Softening for Premium H13 at the different test temperatures and strain ranges.
In paper III it was concluded that the tool materials soften during use and it was also determined that the microstructure has a great influence on the softening rate. This study aim at clarifying what occurs in the microstructure and what the decrease in properties originates from. In paper I it was found that the crack nucleation occurs at very early stage in the tool life, and it is well known that prior to crack nucleation the material experiences substructural and microstructural changes, which cause permanent damage and nucleation [6]. To delay the crack initiation and to enhance the production life a better understanding of the microstructural changes are needed.

X-ray line broadening and TEM observations were used to evaluate the microstructural changes in the tools from paper I and in the specimens from paper II and III. By using X-ray line broadening analysis it is possible to measure the amount of disturbance, which prevails locally in the matrix. The disturbance primarily comes from dislocations, but is also contributed by alloying elements and precipitates, and is expressed as microstrain of the original matrix. The microstrain can change when the material is loaded upon, but can also change when the material is subjected to an increased temperature and is called recovery. To evaluate if the change in microstrain comes from thermal recovery or load, measurements on specimens subjected to increased temperatures and on specimens subjected to a load at room temperature were performed.

From the X-ray measurements on specimens, which were subjected to an increased temperature for 2 hours it was found that the microstrains in DIEVAR decreased at a lower temperature than for QRO 90, Fig. 23(a). The microstrain decrease was lower in the heat treatment condition TA 1100 °C than in TA 1020 °C. Similar results were found when the specimens were subjected to a constant temperature of 600 °C, but for longer times, Fig. 23(b). The microstrains were found to be closely connected to the hardness, Fig. 24. Measurements of tensile specimen at room temperature showed that no microstrain decrease had occurred.
Fig. 23. Microstrain measurements showing the temperature effect in QRO 90 and DIEVAR heat treated with austenitizing temperatures (TA) 1020 and 1100 °C at (a) different temperatures for 2h and (b) at different times at 600 °C.

Fig. 24. a) Microstrain and hardness vs. tempering temperature and b) microstrain and domain size vs. hardness.

If the microstrain decrease from a constant temperature at 600 °C is compared with the isothermal fatigue results at 600 °C it can be seen that the DIEVAR specimens has experienced a greater microstrain decrease than it would have if it was only subjected to the temperature, Fig. 25(a). However, this was not the case for QRO 90, where the softening was of the same amount as from the tempering. The critical temperature for the microstructure to turn unstable is 600 °C for DIEVAR but a little higher for QRO 90, Fig. 23. Therefore DIEVAR experiences a greater decrease of microstrain, which is not only produced from the temperature but also facilitated by stress.
The strength of the material comes from a complex interaction between dislocations-dislocations, dislocations-carbides, dislocations-alloying elements and the martensitic phase as the supersaturated matrix structure. A high dislocation density is produced during martensitic hardening of the hot-work tool steels, which are created during the diffusionless transformation [11]. The sub-structure and amount of the dislocations are believed to be an important contribution to the strength and it is vital that it can be maintained during use. Observations of an isothermally fatigued premium H13 specimen has shown that the microstrains rapidly decreases, just as the stress amplitude decreases, the first load cycle, Fig. 26. The decrease in microstrain and the initial softening at test temperatures of 600 °C was found to originate from a rapid dislocation decrease. Similar results, i.e. a rapid decrease in microstrains the first load cycles until stability is reached, were found for premium H13 and the different heat treatments of DIEVAR when they were subjected to thermal fatigue loading, Fig. 27(a). Even if the various heat treatments of DIEVAR and the premium H13 material had different initial microstrain levels they had approximately the same level after only 10 cycles. This behaviour was also confirmed by tests performed on a machined and on an electro polished premium H13 specimen, where it was found that even if the machined surface initially had a higher amount of microstrains the level was approximately the same after 12 thermal cycles, Fig 27(b).
The initial rapid decrease in microstrains was further analysed with TEM, where a H13 Supreme specimen subjected to 70 isothermal fatigue cycles at a temperature of 600 °C was analysed. Thin foils were prepared from the same specimen, from the waste where deformation had occurred and from the butt where no deformation had occurred. These two locations were observed and compared with each other. In general, it was found that the hardened and tempered martensite contained a high dislocation density and the substructure was formed from the martensite lath. However, no cell structure formation was found. The average size of the martensite lath was found to be 350±100 nm in the Fig. 28(a). Overall, the only characteristic feature of the deformed material found was a few martensite laths with low dislocation density in the interior, Fig. 28(b).
Fig. 28. TEM photograph of the deformed H13 Supreme material showing (a) subgrains as formed by the original martensite lath structure and (b) a large subgrain with low dislocation density in the interior.
TOOL MATERIAL MODELLING (PAPER V & VI)

Numerical simulation of materials behaviour during use has the last twenty years become highly interesting, for many reasons. One of the major reasons is to avoid expensive and time-consuming tests of materials in new and old operation dies by simulating a tool materials performance and life length. However, to simulate materials behaviour a great amount of both experimental and numerical work needs to be done. In paper VI an attempt to simulate the hot-work tool steel behaviour during a specific hot forging was made. A failure criterion based on real life observations worked as a base for the strain amplitudes. The material was tested at several temperatures within the range of the temperatures that the die was exposed to, since the material behaves different at different temperatures. From the experimental work it was then possible to identify the material model parameters needed to simulate the tool behaviour. The parameters used in the material model was then further analysed, with regard to the microstructure, in paper V. But, in paper V the parameters were tested at temperatures close to or above tempering under die-casting conditions, since in many die-casting processes the surface temperature reaches above the materials tempering temperature [1-3].

The model used was an elasto-plastic, non-linear kinematic and isotropic hardening model. Originally the non-linear kinematic part in the model was formed by Armstrong and Fredrick and published 1966 [14]. Then Marquis [15] developed the isotropic part, which takes account for the cyclic softening/hardening, in 1979, and finally, a strain memory variable was introduced in 1979 by Chaboche [15]. Chaboche has further developed the model and has introduced several of other parameters [15]. The model is built on an elasto-visco-plastic behaviour assumption where the stress and strain are partitioned [15-17], i.e. the strain is partitioned into elastic and plastic part and the stress is divided into a kinematic, isotropic and elastic limit part. Sometimes other parameters are also added to the model to compensate for other effects such as ratchetting etc. [15, 18]. The isotropic part describes the change in size of the yield surface, and corresponds to the materials strength due the number of blocked dislocations, which in general,
depends on dislocation structure and density and/or carbide morphology. The kinematic part illustrate the movement of the yield surface and describe the directional stress fields due to dislocation pile-ups at obstacles e.g. precipitates and grain boundaries.

The isotropic parameters are identified from the stress amplitude decrease (or increase) with the number of cycles and was found to be strongly effected by the materials different heat treatments as can be seen for DIEVAR in Fig. 29(a) (where the calculated values also are shown). It was also found that the softening consists of three stages. The primary stage consists of a rapid initial softening, and is described by two parameters and represents approximately the first 3-600 cycles in Fig. 29(a). Usually, the primary rapid initial softening is explained by rearrangement of the tangled dislocations into cell structures with a strongly reduced dislocation density [17, 19], but the observations in paper IV showed that no cell structure was found. An obvious connection between the microstructure in the secondary softening parameters was found, and the carbide morphology was believed to play an important role.

The kinematic parameters are either identified from a tensile test or from the first quarter cycle of an isothermal fatigue test i.e. during the initial load increase. In paper V and VI the kinematic part was identified under the assumption that the isotropic variable is zero during the first load cycle, Fig. 29(b), and is expressed with an exponential function. The two kinematic hardening parameters are identified as the strain-hardening level and the strain-hardening rate. It was found that the strain-hardening level is the highest when using intermediate austenitizing temperature (1060-1100 °C). This indicates that there is an optimal combination of grain size and precipitate distribution affecting the strain hardening level. From paper IV it was found that initially there was a difference in microstrain between different heat treatment conditions, and a distinction between the various heat treatment initial softening rate parameters could be found. Also, it should be noted, that the present model used the initial strain-hardening behaviour to determine the material parameters during the entire simulation. However, it is recognised that it may change during a test, giving additional effects on the secondary softening.
Fig. 29. Experimental and calculated curves used to identify the (a) isotropic and (b) kinematic parameters (paper V).

From the results it was found that when the tool materials behaviour was simulated for the hot forging condition [paper VI] the FE-model could represent the overall behaviour of the experiments considering the two stages of softening and the kinematic hardening within the load cycles. The simulation of the total strain at 200 and 600 °C, were in general in good agreement, Fig. 30(a), but the plastic deformation, Fig. 30(b), were not perfectly simulated.

Fig. 30. Experimental and simulated values for premium H13 at (a) 200 and 600 °C showing the total strain and (b) at 600 °C showing cycle 2 and 500 (paper VI).

When a similar numerical simulation was performed for the die-casting conditions it was confirmed that a good simulations could be found in symmetrical load conditions, Fig. 31, but not in a non-symmetrical condition, Fig. 32. To determine why the simulation did not work properly for the non-symmetrical load condition a test under non-symmetrical forging conditions were performed. It was found that under these conditions (larger total strain) the model worked satisfactorily even under non-symmetrical load conditions, Fig. 33. It was suggested
in paper V that the reason for the error in the simulation of the non-symmetrical die casting test was because the Baushinger and shakedown effect, during the first load cycle, was so proportionally large that they could not be captured as the model was constructed.

Fig. 31. Total strain (-0.3/0.3 %) fatigue test at 600 °C of DIEVAR with Tₐ 1020 °C showing (a) experimental and (b) simulated stress-plastic strain loops 1, 500, 5000 and 10000.

Fig. 32. Total strain (-0.4/0.2 %) fatigue test at 600 °C of QRO 90 with Tₐ 1020 °C showing (a) experimental and (b) simulated stress-plastic strain loops 1, 500, 5000 and 10000.

Fig. 33. Total strain (-0.5/1 %) fatigue test at 500 °C of DIEVAR with Tₐ 1020 °C showing (a) experimental and (b) simulated stress-plastic strain loops 1, 5, 10, 50, 100, 256, 500.
11 PRACTICAL IMPLICATION

From investigations of the different hot forming tools it can be concluded that thermal fatigue is a major failure mechanism. Other studies have also shown that more than 80% of the hot-work dies fail by crack initiation caused by thermal fatigue [20]. It is well known that in order to avoid thermal fatigue the die material should have low coefficient of thermal expansion, high thermal conductivity, high hot yield strength, good temper resistance, high creep strength, adequate ductility and toughness [2, 3]. However, it is impossible to get all these properties optimised in the same material, consequently, some properties must be prioritised. In paper I it was shown that plastic deformation occurs during each thermal cycle and that crack initiation takes place at less than 1% of the tools lifetime. The tool, therefore, spends the majority of its life resisting crack propagation. To increase the tools lifetime one could either increase the number of load cycles to crack initiation and/or increase the resistance to propagation. One way to increase the number of cycles to crack initiation is to minimise the plastic deformation in each cycle, which can be achieved by for example increasing the hot yield strength. From the results in Table 3 it is found that, for the same hardness, the different materials and heat treatments have about the same yield strength. However, they still have a considerable difference in high temperature fatigue life, and most likely crack initiation resistance, Fig. 18 and 21(c), which can be explained by their different softening rates, i.e. resistance to yield strength decrease. From the isothermal fatigue results, Fig. 21, it was found that the heat treatment condition with an intermediate austenitizing temperature (1060-1100 °C) had the strongest resistance to softening, i.e. the material with the most stable microstructure also had the strongest softening resistance. Similar conclusions can also be drawn if the different test materials are compared with each other, where QRO 90 has greater softening resistance and also a longer fatigue life Fig. 21(c). From characterisation of the different materials and conditions it is found that the materials and conditions with greatest isothermal and thermal fatigue resistance have maximised temper resistance, Fig. 3, and lower ductility and toughness, Fig. 5. Therefore, at
high temperature conditions (above the tool materials tempering temperature), which prevails for most hot forming techniques, microstructural stability must be prioritised. By looking at the characterisation results it seems as if it is better to increase the microstructural stability by the choice of material than to increase the austenitizing temperature. Since, QRO 90 with the lowest austenitizing treatment still have higher fatigue resistance, Fig. 21(d), temper resistance, Fig. 3(b) and higher ductility and toughness, Fig. 5, than for the other materials with increased austenitizing temperature.

For tools used at temperatures below the tool materials tempering temperature, the microstructural stability is not as important. In paper IV it was found that no change in microstrain had occurred after 10,000 thermal cycles with a maximum temperature of 600 °C. But, cracking were still found in the specimens’ surface, Fig. 19, even if the material surface did not soften, due to the thermal loads. At these conditions microstructural stability is not as important, and the life of the tool could probably be increased if the hot yield strength or the ductility and toughness were improved. Nevertheless, in many hot forming techniques there is a risk of gross fracture therefore an increase in the materials ductility and toughness would be a better choice.

The strength of a material is associated with resistance to slip and dislocation motion and in a martensite material this is attributed to the deformed crystal structure. There are also other important contributing factors to the strength such as carbide precipitates and dislocation density and structure. However, in paper IV it was shown that the initially different amount of dislocation densities in the different materials and heat treatment conditions hardly had any effect on the softening resistance, Fig. 27. But, it was found that the most important property in order to resist softening is temper resistance, which mainly is controlled by the carbide morphology. In the heat treatment conditions where a higher austenitizing temperature is used more of the primary carbides are dissolved into the austenite, which not only makes the martensite more saturated with alloying elements, but also in the following tempering the amount of small stable secondary carbides will increase. If the carbides between the different materials are compared it can first of all be seen that the majority of carbides in the material are chromium rich carbides such as M₆C₃ and M₂₃C₆, Fig. 4(b), which easily coalesce and coarsen. However, QRO 90 has lower amount of chromium and higher amount of vanadium, Table. 1, thus it most likely contains carbides that are more stable [11, 21].
Conclusively the increased softening resistance in QRO 90 primarily comes from the more stable carbides.

The demand on more efficient hot-work industry increases and since much of the efficiency derives from the tool and its lifetime, the “Computational Engineering” becomes extremely important. Component and die design are closely linked, where Computer Aided Design (CAD) and Finite Element (FE) stress calculations are potential means to minimise delay and to increase the tool lifetime. Models of the tool steels behaviour is needed in the FE calculations and the results from the numerical simulation in this study show that it is possible to use the non-linear kinematic and isotropic hardening model to simulate the behaviour of a tool during use. However, the model has also shown that it is difficult to simulate the tool steel behaviour at temperatures around tempering and that more work needs to be done in this area. For example, a greater effort to separate the kinematic and isotropic contributions must be done, which is difficult since the material behaviour is time depended at temperatures around tempering. Also, other parameters such as ratchetting should be added.
12 CONCLUSIONS

The following general conclusions are based on the conditions that prevails for the aluminium die casting and hot forging condition investigated in this study.

- Thermal fatigue is found to be a major damage mechanism, and the cracking is initiated at less than 1% of the tools life. The tools spend the majority of its life resisting crack propagation.
- Chromium hot-work tool steels soften at high temperature fatigue loading, and the softening can be divided into three stages: a primary, a secondary and final fracture stage.
- In the primary stage softening at test temperatures close to tempering a dislocation density decrease of 1/3 takes place, which primarily associated with dislocation rearrangement and annihilation. It was also found that, at these conditions, the initial microstrain level had only a minor effect on the softening resistance, since the different heat treatment conditions ended up with approximately the same level after only a few load cycles.
- The second stage softening rate was found to be strongly connected with the temper resistance, which, in general, is controlled by the carbide morphology, and it was found that the intermediate heat treatment conditions had the most stable carbide morphology.
- It was found that the thermal fatigue and the isothermal fatigue life could be increased if the second stage softening rate could be decreased, for example by different heat treatments.
- A kinematic and isotropic hardening model can be used to simulate the hot-work tool steels behaviour in industrial applications, which can lower design and development costs by excluding expensive experiments and tooling.
13 REFERENCES

13. I. Schruft, *Comparison of properties and characteristics of hot-work tool steels X38CrMoV5 1, X40CrMoV5 1, X32CrMoV3 3, X38CrMoV5 3*, Technische Berischte/Thyssen-Edelstahl, 1990, p.32-44.


Chromium martensitic hot-work tool steels

Chromium martensitic hot-work tool steel (AISI H13) was developed to endure the severe conditions of high temperature metal forming operations such as die casting, hot rolling, extrusion and hot forging. The mechanical properties are high and strongly connected to the microstructure and have been improved over the years by alloying and heat treatment. Damages still occur and one of the most common failure mechanisms is thermal fatigue.

In this thesis the thermal fatigue damage on hot forming tools has been studied. Several types of hot work tools steels have been experimentally tested and the microstructural changes during thermal fatigue have been evaluated. The tool material behaviour has also been simulated to support the integration of die design, tool steel properties and use.

The general aim of this thesis is to increase the knowledge of the chromium martensitic hot-work tool steel damage, performance and microstructure.