On Tool Failure in Die Casting

BY

ANDERS PERSSON
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Abstract

Die casting is a very cost-efficient method of forming thin-walled and complex near net-shaped products with close geometric tolerances and good surface finish. A permanent die tool is used to make large quantities of identical products. The performance and tool life are limited by several mechanisms, e.g. thermal fatigue cracking, erosion, and corrosion. To develop new and more resistant tool materials for die casting detailed knowledge of the actual casting conditions and the tool failure mechanisms are essential. This thesis contributes to an increased knowledge of tool failure in die casting by investigating and simulating actual casting conditions and tool failure mechanisms.

A method to record the temperature fluctuations in a cavity insert during actual brass die casting was developed, and details of the temperature conditions were obtained. Also, a test method based on cyclic induction heating and internal cooling of hollow cylindrical test rods was developed, where the surface strain during thermal cycling could be measured. This method reproduced the characteristic type of surface cracking observed on die casting tools, and proved to give information of the strains and stresses behind the fatigue failure.

In actual die casting, the dominant tool failure mechanism is thermal fatigue cracking. The formation of the cracks is associated to accumulation of the local plastic strain that occurs during each casting cycle. Initial crack growth is facilitated by oxidation of the crack surfaces, and proceeded growth is facilitated by this oxidation in combination with crack filling of cast material, and by softening of the tool material. In addition, local enrichment of Pb at the crack front from the cast alloy melt was also observed to promote the crack growth in die casting of brass.

In an investigation of thermal fatigue of two hot work tool steels, quenched and tempered to different conditions, it was found that low-cycle fatigue occurs, although the estimated tensile stress never exceed the initial yield strength of the steel. The reason is a gradual softening of the steel during the thermal cycling, and the presence of stress raising defects. The resistance against thermal cracking improves with initial tool steel hard ness, because any initial ranking in hardness among the steels is unaffected by the thermal cycling.

Another investigation on a selection of surface engineered tool steels, including common diffusion treatments, PVD coatings and combinations of these, showed that surface engineering generally reduce the resistance against thermal cracking as compared to untreated references, since the engineering processes influence negatively on the mechanical properties of the hot work tool steels.

Finally, corrosion tests of CrN PVD-coated tool steels by exposing them to molten aluminium revealed the mechanisms of initiation and progress of liquid metal corrosion of this material combination, and that the corrosion resistance improves with the CrN coating thickness.

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List of enclosed papers

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


II A. Persson, S. Hogmark, J. Bergström, “Failure modes in field-tested brass die casting dies”, In manuscript.


V A. Persson, “A strain-based approach to crack growth and thermal fatigue life of hot work tool steels”, In manuscript.


The author’s contribution to the papers

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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 Tribomaterials Group, Ångström Laboratory, Uppsala University. The financial and material support from the participating companies in the research project Uddeholm Tooling AB, Tour & Andersson AB, and Bodycote Heat Treatment AB is gratefully acknowledged. Furthermore, the financial support from the Swedish Knowledge Foundation is also acknowledged.

First of all, I would like to extend my gratitude to Prof. Sture Hogmark at the Tribomaterials Group, Uppsala University, for being an excellent supervisor who always give exceptional guidance and response. My highest gratitude goes also to Prof. Jens Bergström at the Department of Materials Engineering, Karlstad University, for organising the research project and also for valuable discussions and tutoring through the years. Prof. Jens Bergström was also one of those who inspired me to choose a research career on materials science, and later on watched my first stumbling steps on the way of science. I would also like to thank my past and present colleagues and the participating personnel at the above mentioned companies for, in various ways, contributing to the success of this work.

Finally, I am very grateful to my family and, last but not least, my fiancée Linda.

Anders Persson, Karlstad, April 2003
1 Introduction

Die casting is a very cost-efficient method of forming thin-walled and complex near net-shaped products with close geometric tolerances and good surface finish. Low melting point alloys based on aluminium, zinc, magnesium, and copper are frequently used. Many different types of products are manufactured by die casting, such as engine blocks, cylinder head covers, valves, pipe couplings, etc. and other components for the automotive industry and for heating, ventilation and sanitary installations, etc.

In die casting, large numbers of identical products are produced using one die, which is necessary, since the die is very expensive. Any kind of tool failure that causes rejection of castings and extra tool maintenance may severely increase the production costs.

The overall aim of this thesis is to aid in the development of better tool materials for die casting. Die casting tools are exposed to severe thermal, mechanical and chemical conditions during each casting cycle. The considerably shorter tool lifetimes when casting aluminium and copper alloys as compared to zinc alloys indicates that melting temperature of the cast alloy is crucial, see Table 1.

Table 1. Typical temperature of the casting alloy melt and normal lifetimes of some die components.

<table>
<thead>
<tr>
<th>Casting alloy</th>
<th>Melt temperature [°C]</th>
<th>Cavity insert [castings×10^3]</th>
<th>Core [castings×10^3]</th>
</tr>
</thead>
<tbody>
<tr>
<td>Zinc</td>
<td>430</td>
<td>500-1,000</td>
<td>500-1,000</td>
</tr>
<tr>
<td>Aluminium</td>
<td>700</td>
<td>100-250</td>
<td>50-150</td>
</tr>
<tr>
<td>Copper (brass)</td>
<td>970</td>
<td>10-100</td>
<td>3-20</td>
</tr>
</tbody>
</table>

The service life of tools for aluminium and brass die casting is limited because of several tool failure mechanisms, of which the most important are thermal fatigue cracking, erosion, corrosion, local adherence of the casting alloy to the tool, and gross fracture. Thermal fatigue cracking dominates in die casting of aluminium and brass alloys, since the temperature of the melt is high. Corrosive attack on the tool surface originating from the aggressive nature of molten aluminium is an important failure mechanism when casting
this alloy. Die casting of these two casting alloys, and tool failure due to thermal fatigue cracking and liquid aluminium corrosion is also the focus for this doctoral thesis.

Finally, to develop new and more resistant tool materials for die casting applications, detailed knowledge of the actual casting conditions and the tool failure mechanisms are essential.

In this thesis, the following main aspects were selected to contribute to an increased knowledge of tool failure in die casting:

- Development of an experimental method to determine the tool temperature fluctuations during actual brass die casting.
- Identification and evaluation of the tool failure mechanisms after actual brass die casting.
- Development of a test method for experimental simulation of thermal fatigue cracking of tool materials.
- Applying the experimental thermal fatigue test method for evaluation and detailed studies of surface cracking of tool materials and surface engineered materials.
- Evaluation of protective surface coatings for reduced corrosive wear when exposed to liquid aluminium.
2 Die casting

A distinguishing characteristic of the die casting process is that the liquid metal is forced by the application of pressure to flow with high velocity during injection and completely and rapidly fill an internally cooled mould, typically within the order of milliseconds [1-3]. The high melt velocity during injection and the continuous internal cooling of the tool during the process allows production of thin-walled and complex near net-shaped cast products at high manufacturing rates, typically of the order of 100 castings per hour per cavity. The complexity of the castings and the manufacturing rate are considerably beyond those provided by the permanent mould (gravity) casting process.

2.1 Die casting machines

There are two principal types of die casting machines, either classified as hot chamber or cold chamber machines, which fundamentally only vary with respect to the method of molten metal injection into the die cavity [1-3]. In the hot chamber machine process, the liquid metal injection mechanism is submerged in the molten metal, and the cylinder is automatically filled with metal prior to each casting operation when the injection plunger rod is withdrawn. This process is typically applied for low melting point casting alloys, which cause the minimum of attack on the injection system material during the contact with the liquid metal (e.g. zinc and magnesium alloys).

In the cold chamber machine process, the injection mechanism is separated from the molten metal, and the cylinder is filled with metal prior to each casting operation using a ladle. A schematic of a cold chamber machine in charging position is shown in Fig. 1. The cold chamber process minimises the liquid metal exposure of the injection system components, and it is normally applied for casting alloys with higher melting points, e.g. aluminium and copper alloys. Again, these casting alloys are also the focus for this thesis.
2.2 The die casting die

A die casting die is a complex tool usually composed of two die halves, named the cover (or fixed) and ejector die half, respectively [1-3], cp. Fig. 1. The cover die half is fixed to the stationary platen of the machine, and the ejector die half is secured to the moveable ejector platen. The die cavity, in which the castings are formed, is constructed by cavity inserts fixed to each of the two die halves, stationary and/or moveable cores, as well as other moveable parts. A cavity insert and a core for die casting of pipe couplings is shown in Fig. 2. The die includes also ejector pins, which are used to eject the castings from the tool when the die opens.

A die can be constructed as a single or multiple-cavity die to produce single or multiple castings of identical shape during each casting operation cycle.
2.3 The die casting operation

In a die casting operation cycle, the die halves are closed and the moveable tool components are moved into position and locked by the clamping system of the machine [1-3], cp. Fig. 1. Prior to casting aluminium and copper alloys, the die is normally preheated to a temperature within the range of 250-300 °C and 300-350 °C, respectively, to reduce the thermal shock when the liquid metal makes contact with the tool material.

Thereafter, the molten metal, which is filled in the chamber, is forced by the plunger rod to flow with high velocity during injection and completely and rapidly fill the die cavity. The metal pressure during the injection is high, the peak of which for aluminium and copper based alloys is typically within the range of about 20-70 MPa and 40-100 MPa, respectively, or more (cold chamber process). The locking force needed to keep the die closed during each casting cycle is related to the peak injection pressure and the projected area of the tool, and it is also the usual denotation of the size of the die casting machine. Machines with locking forces within 450 kN-35 MN are available. For aluminium alloys, the entrance velocity of the melt is usually 20-60 m/s and the melt temperature is approximately 700 °C, whereas those for copper alloys is about 1-10 m/s and 970 °C.

The molten metal in the cavity solidifies rapidly through the internal cooling of the die. When the casting has solidified, the die halves and the moveable tool components are unlocked and the mobile parts withdrawn. Thereafter, the die halves are opened and the casting ejected. Subsequently, the tool surfaces may be externally cooled and lubricated by spraying to prevent adherence of the next casting. The complete die casting operation cycle is very rapid, and the process allows high manufacturing rates of the order of 100 castings per hour per cavity.
2.4 Aluminium and copper based die casting alloys

Aluminium alloys in the aluminium-silicon systems AA A3xx.x and A4xx.x are the most important aluminium compositions for die casting [1,4]. The principal alloying element in both systems is silicon, and in the A3xx.x group is also copper and/or magnesium added as alloying elements. Silicon improves the fluidity and castability of aluminium alloys considerably, and copper and magnesium improves strength and hardness through precipitation of intermetallic compounds. Fluidity is a measure of the distance to which liquid metal will flow before it solidifies, and it is one of the factors that determine the ability of the molten alloy to completely fill the die cavity. Castability is the ability of the molten metal to reproduce fine detail on the surface. Additions of iron in aluminium die casting alloys reduce the tendency for adherence of the casting alloy to the die during casting ejection. Two common aluminium die casting alloys are A380.0 and A413.0.

Copper alloys in the classes of yellow brasses, silicon brasses, and silicon bronzes are common copper based alloys for die casting [1,4,5]. The casting characteristics of copper are improved by adding alloying elements, such as zinc, silicon, tin, beryllium, nickel, and chromium. Additions of lead in copper alloys improve the machinability. Some common copper based die casting alloys of the above classes are UNS C85800, C87900, C87500, and C87800.
3 Materials and surface engineering for die casting dies

3.1 Tool materials

The essential properties of tool materials for die casting applications are high levels of hot hardness (or hot yield strength), temper resistance, creep strength, ductility, toughness, oxidation resistance, along with low thermal expansion coefficient, and high thermal conductivity [2,3,6]. Hot work tool steels offer a wide selection of candidates that fit many of those property requirements. This is also the reason why dies for die casting of aluminium and copper alloys most frequently are made of chromium and tungsten hot work tool steels, e.g. AISI H11, H13, H21, or H22. They are used in quenched and tempered condition often within a hardness range of about 360-520 HV, depending on die component and steel grade [2]. The range in chemical compositions of some hot work tool steel grades, their typical hardening and tempering temperature range, and usual working hardness, along with the nominal chemical compositions, typical hardening and tempering temperature range, and hardness range of the hot work tool steel grades used in this thesis are given Table 2.

3.1.1 Hot work tool steels

Hot work tool steels are classified as either chromium (H10-H19), tungsten (H21-H26), or molybdenum (H42-H43) hot work tool steel grades according to the major principal alloying element [7-9]. They are usually alloyed with low or medium carbon content levels (0.35-0.45 wt. %) to promote toughness, and additions of the strong carbide forming elements such as chromium, tungsten, molybdenum, and vanadium to promote high temper resistance and hot hardness, cp. Table 2. The strong transition carbide forming elements are distributed in the hardened and tempered martensitic microstructure as primary alloy carbides retained after hardening, fine
Table 2. Range in chemical composition [wt. %] of frequently used hot work tool steel grades, typical hardening (H) and tempering (T) temperature range, and usual working hardness [HV] [2,7,8], and nominal chemical compositions, typical hardening and tempering temperature range, and hardness range of the Uddeholm grades QRO 90 Supreme, Orvar Supreme (premium H13 grade), and Hotvar used in this thesis.

<table>
<thead>
<tr>
<th>Material</th>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>Cr</th>
<th>Mo</th>
<th>V</th>
<th>W</th>
<th>Ni</th>
<th>H</th>
<th>T</th>
<th>Hard.</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>°C</td>
<td>°C</td>
<td>°C</td>
<td>°C</td>
<td>°C</td>
<td>°C</td>
<td>°C</td>
<td>°C</td>
<td>°C</td>
<td>°C</td>
<td>HV</td>
</tr>
<tr>
<td>AISI</td>
<td>0.33-0.80</td>
<td>0.20-4.75</td>
<td>1.10-0.30</td>
<td>---</td>
<td>max</td>
<td>995-540</td>
<td>540-370</td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>H11</td>
<td>0.43</td>
<td>1.20</td>
<td>0.50</td>
<td>5.50</td>
<td>1.60</td>
<td>0.60</td>
<td>0.30</td>
<td>1025</td>
<td>650</td>
<td>620</td>
<td></td>
</tr>
<tr>
<td>AISI</td>
<td>0.32-0.80</td>
<td>0.20-4.75</td>
<td>1.10-0.80</td>
<td>---</td>
<td>max</td>
<td>995-540</td>
<td>390</td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>H13</td>
<td>0.45</td>
<td>1.20</td>
<td>0.50</td>
<td>5.50</td>
<td>1.75</td>
<td>1.20</td>
<td>0.30</td>
<td>1040</td>
<td>650</td>
<td>580</td>
<td></td>
</tr>
<tr>
<td>AISI</td>
<td>0.26-0.15</td>
<td>0.15-3.00</td>
<td>---</td>
<td>0.30-8.50</td>
<td>max</td>
<td>1095</td>
<td>595</td>
<td>390</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>H21</td>
<td>0.36</td>
<td>0.50</td>
<td>0.40</td>
<td>3.75</td>
<td>0.60</td>
<td>10.00</td>
<td>0.30</td>
<td>1205</td>
<td>675</td>
<td>620</td>
<td></td>
</tr>
<tr>
<td>AISI</td>
<td>0.30-0.15</td>
<td>0.15-1.75</td>
<td>---</td>
<td>0.25-10.00</td>
<td>max</td>
<td>1095</td>
<td>595</td>
<td>350</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>H22</td>
<td>0.40</td>
<td>0.40</td>
<td>0.40</td>
<td>3.75</td>
<td>0.50</td>
<td>11.75</td>
<td>0.30</td>
<td>1205</td>
<td>675</td>
<td>600</td>
<td></td>
</tr>
<tr>
<td>QRO 90</td>
<td>0.38</td>
<td>0.30</td>
<td>0.75</td>
<td>2.60</td>
<td>2.25</td>
<td>0.90</td>
<td>---</td>
<td>---</td>
<td>1020</td>
<td>615</td>
<td>430</td>
</tr>
<tr>
<td>Orvar</td>
<td>0.39</td>
<td>1.00</td>
<td>0.40</td>
<td>5.20</td>
<td>1.40</td>
<td>0.90</td>
<td>---</td>
<td>---</td>
<td>1025</td>
<td>600</td>
<td>450</td>
</tr>
<tr>
<td>Hotvar</td>
<td>0.55</td>
<td>1.00</td>
<td>0.75</td>
<td>2.60</td>
<td>2.25</td>
<td>0.85</td>
<td>---</td>
<td>---</td>
<td>1050</td>
<td>575</td>
<td>600</td>
</tr>
</tbody>
</table>

Precipitated alloy carbides formed during tempering, and as alloying elements in the tempered martensitic structure.

A number of various types of alloy carbides with different characteristics have been identified in tool steels, M1C, M2C3, M23C6, M6C, M7C and MC, where M denotes the metal content of the carbide. The carbide type and volume fraction is dependent upon the carbide-forming element alloy level, carbon content, and heat treatment. An increase in the carbide-forming element alloy content and the carbon level is associated with a higher alloy carbide fraction level and, therefore, higher tool steel hardness and wear resistance. In addition, higher carbide hardness also contributes to higher tool steel hardness and wear resistance. However, a large amount of carbides in the microstructure is associated with reduced tool steel toughness. The stability of the alloy carbides in the microstructure at high temperature exposure is the origin for the high temper resistance and hot hardness of tool steels.

The temper resistance and hot strength of tool steels is improved by alloying with higher contents of the carbide-forming elements tungsten and molybdenum to produce large volume fractions of thermally stable carbides in the microstructure. In addition, alloying with vanadium improves the resistance against high temperature wear, since the vanadium carbide is the
hardest of all carbides present in tool steels. The oxidation resistance of tool steels is improved by alloying with higher levels of silicon.

The tungsten and molybdenum hot work tool steels grades have higher temper resistance and hot hardness than the chromium hot work tool steel grades, since they are alloyed to give large volume fractions of thermally stable alloy carbides in the microstructure. However, the large amount of carbides in the microstructure is associated with reduced tool steel toughness. The higher hot hardness and temper resistance of the tungsten hot work tool steels is the reason why they commonly are selected in brass die casting applications [5,10,11]. The temper resistance and relevant mechanical properties at various temperatures of some hot work tool steels used in this thesis are given in Fig. 3.

Fig. 3. Nominal properties of some hot work tool steels, hardened and tempered to various conditions, used in this thesis. Temper resistance as hardness vs. holding time at various temperatures (a), ultimate tensile strength as $R_m$ (b), tensile yield strength as $R_{p0.2}$ (c), ductility as reduction of area at tensile fracture (d), and modulus of elasticity (e) vs. temperature.
Surface engineering

Surface engineering of ceramic materials, which typically are characterised by high hardness, wear resistance, and chemical inertness, have successfully been applied on tool materials in die casting applications to improve the performance and tool life [12-16]. The resulting tool surface layer properties have proven to improve the resistance against erosive and corrosive attack originating from the molten metal flow, and in some cases also the surface cracking performance due to thermal fatigue [12-21].

In this thesis, various types of surface engineering were used, including nitriding, boriding, Toyota diffusion, and physical vapour deposition (PVD) of coatings as single-layers or on top of a nitrided layer (duplex-treatment). Brief descriptions of the engineering processes are given below.

In the PVD process, solid metal is vaporised to subsequently form a surface layer on top of the tool material by condensation from the vapour phase [22]. Thin films of ceramic compounds, e.g. CrC, CrN, TiN and TiAlN, are typically deposited. The deposition temperature is about 200-600 °C.

Nitriding is a surface modification process that involves diffusion of nitrogen into the surface material [23,24]. In the nitriding process, nitrogen is added to the surface material in a gas, liquid (salt-bath), or plasma (ion) environment, typically at 480-590 °C, 510-565 °C, and 340-565 °C, respectively. The surface layer of a nitrided steel consists of a diffusion zone, made of the original microstructure with additions of nitride precipitations and nitrogen in solid solution, with or without a compound layer of Fe2.3N (ε-phase) and Fe3N (γ'-phase) as the top layer. The compound layer is known as the white layer.

In the boriding process, steel is kept in contact with a reactive boron containing compound at high temperature, typically 840-1050 °C, to produce a surface layer of borides [24,25]. Initially, an iron boride layer is formed at the surface and grow subsequently in thickness due to diffusion of boron into the material. The diffusion of boron into the steel substrate push elements that are not soluble in the boride layer (e.g. carbon and silicon) into the substrate, and cause an enrichment of these elements in a zone below the boride layer. A single-phase layer of Fe2B, with a characteristic sawtooth like morphology, is more desirable than a double-phase layer with the more brittle FeB compound. Tools are often hardened and tempered after the boriding treatment.

In the Toyota diffusion process (TDP), steel is kept in contact with carbide-forming elements such as vanadium, niobium, and chromium at high temperature, typically 800-1250 °C, to produce surface layer of carbides [24,25]. Initially, carbon in the steel substrate combines with the carbide-
forming element and form a carbide layer on the surface. Thereafter, the carbide layer grow thicker due to further reactions at the carbide layer interface as a result of diffusion of carbon from the substrate to the surface. During the process, the carbide-forming element vanadium or chromium diffuse into the substrate and form a solid solution layer beneath the carbide layer. As an example, the desired carbide layer during TDP to give CrC is of the type Cr$_7$C$_3$ or Cr$_{23}$C$_6$. Tools are usually quenched and tempered after processing.

Example of the characteristics of the surface engineered layers and deposition temperatures used in this thesis are shown in Table 3.

Table 3. Example of characteristics of the surface engineered layers and deposition temperatures used in this thesis.

<table>
<thead>
<tr>
<th></th>
<th>PVD CrN</th>
<th>PVD CrC</th>
<th>PVD TiAlN</th>
<th>Plasma nitriding</th>
<th>Gas nitriding</th>
<th>Boriding</th>
<th>TDP CrC</th>
</tr>
</thead>
<tbody>
<tr>
<td>Coating hardness</td>
<td>2000</td>
<td>1700</td>
<td>2300</td>
<td>900-1000</td>
<td>950-1050</td>
<td>1750</td>
<td>2000</td>
</tr>
<tr>
<td>[HV 25 gf]</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Coating thickness</td>
<td>4-6</td>
<td>4-6</td>
<td>3-6</td>
<td>---</td>
<td>---</td>
<td>---</td>
<td>---</td>
</tr>
<tr>
<td>[µm]</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Diffusion depth</td>
<td>---</td>
<td>---</td>
<td>---</td>
<td>100-200</td>
<td>250-300</td>
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<td>Deposition</td>
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4 Failure mechanisms in die casting dies

The most important tool failure mechanisms that limit the performance and service life of tools in aluminium and brass die casting applications are thermal fatigue cracking, erosion, corrosion, local adherence of the casting alloy to the tool (soldering), and gross fracture [1-3].

Thermal fatigue cracking results from the rapid fluctuations in die surface temperature during the casting process, which may induce stresses high enough to impose an increment of plastic strain in the tool surface during each casting cycle. Surface cracks appear already after a few thousand castings, or even earlier, and are, therefore, formed in the low-cycle fatigue range [26,27]. Oxidation and creep may significantly contribute to cracking [26-30]. Thermal fatigue cracking is often observed as a network of fine cracks on the tool surface. This type of crack pattern is often named heat checking. Another variant of thermal cracks (stress cracks) may be observed as individual and clearly pronounced cracks in areas exposed to local stress concentrations [26]. The formation of thermal cracks may lead to loss of surface material in the form of small fragments (splintering).

Erosion is induced by the high velocity of the incoming melt and is partially aggravated by the presence of solid particles in the molten casting alloy. The erosive damage is primarily seen where the molten metal jet first hits the die surface.

Corrosion damage originates from dissolution of the tool material into the molten metal and formation of intermetallic compounds. For aluminium alloys, the corrosive attack may form craters on the tool surface [31], as well as small protuberances of Fe-Al intermetallic phases [32]. The erosive and corrosive attack from the liquid metal interaction may cause considerable loss of surface material.

Soldering results from the chemical interaction between the die material and the casting alloy during filling and solidification and may be observed as residuals of the casting alloy on the die surface after ejection.

Gross fracture is primarily caused by excessive thermal shocks or mechanical overloading, and leads often to total failure of the tool.

The mentioned failure modes may, ultimately, increase the production costs through extra tool maintenance, rejection of castings, and expensive die failures.
5 Tool temperature conditions during actual brass die casting (Paper I)

The rapid temperature alternations in the tool surface layer, through the cyclic nature of the die casting process, is the origin of thermal fatigue cracking of die casting tools. Detailed knowledge of the thermal cycling during actual die casting is needed to perform realistic and well-controlled imitations of the heat cycling in simulations, and also to increase the understanding of the conditions during which tool failure occurs. To achieve this type of information for brass die casting, an experimental method to determine the tool temperature fluctuations during actual service conditions was developed, see Paper I.

5.1 Field test equipment (Paper I)

A cavity insert was instrumented with four measurement probes to obtain recordings of the temperature fluctuations at different depths below the surface during actual die casting of brass, see Fig. 4. Each of the measurement probes included a cylindrical test specimen (∅ 8 mm) and a K-type (Chromel-Alumel) thermocouple with thin wires (∅ 0.13 mm), spot welded to the back of the specimen, see Fig. 4b. The thin wires enable rapid response of any change in temperature. The temperature from all thermocouples was recorded with a sampling rate of 500 Hz.

The tests were made in a 1.5 MN cold chamber machine during production of brass tube couplings such as that of Fig. 4b. The temperature of the brass melt was 980 °C and the total cycle time 30 s during which the die was closed 10 s and open 20 s. Water at 20 °C was continuously circulated through cooling channels in the tool. In addition, the tool surfaces were lubricated but not intentionally cooled by spraying. The total shot weight of each casting was 1.6 kg, the peak casting pressure 164 MPa, and the entrance velocity of the melt was about 1.5 m/s. Note that the cavity insert was not preheated.

The cavity insert, measurement probes, and test specimens were made of QRO 90 Supreme. Test specimens in quenched and tempered condition as
well as surface engineered by boriding and PVD coating of CrN, respectively, were included, see Paper I.

Fig. 4. Principles of the instrumentation of the cavity insert. (a) Section of the insert showing the location of the four temperature measurement probes (the cylindrical components in the tool cavity). The thickness in mm of the test specimens is indicated. (b) A typical casting and the bottom probe missing in (a) with its thermocouple. (c) Details of the probe. (Paper I)

5.2 Temperature cycling of the die surface (Paper I)

It was found that, starting with a tool of room temperature, it takes less than 20 casting cycles to establish a cyclic equilibrium tool temperature of about 300 °C. Thereafter, when the 980 °C brass alloy melt makes contact with the tool, the tool material is heated within about 0.35 s from around 300 °C to a maximum temperature of around 750 °C at a depth of 0.25 mm, see Fig. 5. Deeper below the tool surface the maximum temperature is lower, and the heating rate is reduced. Until the tool is opened, cooling occurs by heat conduction into the bulk of the tool. Die opening and simultaneous cast ejection give rise to an additional heat loss through irradiation and convection, which naturally is most notable at the superficial depths. No extra cooling effect from the lubricant spraying was observed. Finally, the surface engineered layers from the boriding or CrN treatment did not have any notable influence on the temperature conditions.
In Paper I, the maximum surface temperature was estimated based on the temperature profiles and by approximating the cavity insert as a semi-infinite solid with an initially uniform temperature, which instantaneously is exposed to a constant surface heat flux. The maximum surface temperature and surface heat flux was estimated to approximately 826 °C and about 9.8 MW/m², respectively, cp. Fig. 6. Similarly, corresponding values for aluminium die casting can be estimated to roughly 520 °C and 5 MW/m², respectively, using the temperature profiles obtained by Venkatasamy *et al.* [33]. This reveals the more severe tool temperature conditions in die casting of brass than of aluminium alloys.
6 Failure of brass die casting tools (Paper II)

The significantly higher melt temperature in die casting of brass, the lower entrance velocity of the melt, the chemical properties of the casting alloy, as well as other process parameters clearly separates the casting conditions of brass and aluminium. To increase the knowledge of tool failure in brass die casting, an investigation of failure mechanisms in cavity inserts and cores after actual brass die casting was conducted, see Paper II.

6.1 Field test conditions (Paper II)

Two cavity inserts made of QRO 90 Supreme, and eight cores made of QRO 90 Supreme and Hotvar, respectively, were examined and evaluated with respect to failure mechanisms after actual production of brass valves. All cavity inserts and cores were field tested in hardened and tempered conditions, see Paper II.

The field tests were made in a 2.8 MN cold chamber machine. The temperature of the brass melt was 970 °C and the total cycle time for one casting was 48 s. The entrance velocity of the melt was 1-2 m/s and the maximum metal pressure during injection was approximately 190 MPa. The total shot weight of each casting was 2.1 kg. Water at 20-50 °C was continuously circulated through cooling channels in the cavity insert. Typical tool maintenance, which includes lubrication after each casting and intermittent cleaning by shot-blasting, were applied during the service conditions. Note that the tools were not preheated.

6.2 Tool failure (Paper II)

Investigation of the cavity inserts and cores after use in actual die casting of brass revealed that the dominating failure mechanism was thermal fatigue cracking, see Fig. 7. Typically, the macroscopic surface damage is observed as a network of surface cracks, and as individual and clearly pronounced cracks at sharp corners, see Fig. 7. The former type is often referred to as heat checking and the latter as stress cracks [26]. Naturally, the tool areas
that are exposed to the most critical thermal and/or mechanical stress concentrations, such as sharp corners and protuberances, or parts of the tool which have been more or less surrounded by the melt, show the worst thermal cracking [26,28], cp. Fig. 7. The fact that the cores are exposed to more severe conditions than the inserts was reflected by the substantially deeper thermal crack penetration, see Paper II.

Fig. 7. Typical macroscopic surface damages observed on cavity inserts and cores after their expired tool lifetime of 45-50×10³ and 10-15×10³ castings, respectively. (a) Overview of an insert revealing local severe surface cracking (A) and individual cracks at sharp geometrical corners (B). (b) Overview of a core revealing a severe network of surface cracks, individual cracks at sharp radiuses, and remnants of solidified brass within the cracks (indicated by the arrow). (Paper II)

6.2.1 Thermal cracking (Paper II)

Examination of the cores revealed that the first thermal cracks were formed during the initial 500 brass die casting cycles, see Paper II. This indicates that the temperature alternations in the tool surface layer during the process (recall Fig. 5) are severe enough to impose stresses high enough to cause local accumulation of plastic strains to nucleate and propagate thermal cracks in the tool surface by a low-cycle fatigue process.

It was found that the initial growth of the thermal cracks is associated with an oxidation attack on the crack surfaces to form a Cr and O rich layer in-between the oxides filling the crack and the virtually unaffected tool steel [27,28], cp. Fig. 9. The thermal cracks grow considerably during the casting process and reach several mm in depth while they are typically filled with a mixture of oxides [27], and residuals from the brass casting alloy [28], cp. Figs. 10-12. In addition, this proceeded crack growth occurs with additional oxidation attack on the crack surfaces. The oxides filling the cracks consist mainly of iron oxides [28], but also of zinc oxides and lead oxides [27], cp. Figs. 9-12 and see Paper II.
Fig. 9. Polished cross-section through a typical crack of a worn-out core in high magnification. (a) Light optical microscopy (LOM) image. (b-f) Energy dispersive X-ray spectroscopy (EDS) maps. The element concentration is proportional to the brightness. (Paper II)

Fig. 10. Polished cross-section of a worn-out core revealing typical long and short cracks (LOM). The frames indicate EDS-mapped areas, see below. (Paper II)

Fig. 11. EDS-maps of the area indicated by frame A in Fig. 10. The element concentration is proportional to the brightness. (Paper II)
An interesting finding was that Pb from the brass casting alloy locally is enriched at the crack front, and that there is a local advancement of the thermal crack front in these areas, see Fig. 13. Obviously, the chemical interaction with Pb from the brass alloy and the tool steel is able to locally assist the crack growth.

Crack length measurements after different numbers of castings revealed that the crack length after relatively few casting cycles was suppressed by higher initial tool steel hardness, see Paper II. Subsequently, the differences in crack length among the tools evens out with number of cycles as the surface layer hardness of the steels gradually averages.
7 Simulation of thermal fatigue cracking (Paper III)

Several experimental test techniques to reproduce the type of surface cracking observed on tools in e.g. die casting applications are reported in the literature. The rapid fluctuations in surface temperature responsible for this type of failure is often applied using heating in molten metal and subsequent cooling by water [17] or pressurised air [15], or induction heating followed by water cooling [19,20].

In this thesis, an experimental method to simulate and evaluate thermal fatigue cracking of tool materials with the possibility to obtain recordings of the surface strain during thermal cycling was developed, see Paper III. The test method was applied on tool steels in quenched and tempered as well as surface engineered conditions, see Papers III-VI.

7.1 Thermal fatigue test equipment and test conditions (Paper III)

The characteristic thermal fatigue cracking observed on the surface of die casting tools is reproduced on the surface of a test rod by simulating the rapid alternating temperature conditions typical for this application. The test equipment is based on cyclic induction heating and internal cooling of hollow cylindrical test rods with a diameter of 10 mm, a length of 80 mm and having a 3 mm axial hole for internal cooling. Surface strain measurements through a non-contact laser speckle technique makes it possible to calculate the strains induced in the specimen surface during thermal cycling, see Figs. 14 and 15.

The laser speckle technique is based on the fact that a surface area illuminated by a laser beam displays a characteristic pattern of small visually observed dots. This “speckled pattern” can be used to estimate the surface strains during thermal cycling. In this test equipment, a HeNe laser is used to illuminate a spot (about 1.5 mm) on the diffusely reflecting specimen surface. Two pairs of symmetrically and perpendicularly arranged CCD-array sensors (Figs. 14 and 15) are used to record the movement of the
speckle pattern during thermal cycling with a sampling rate of 500 Hz. From these movement recordings, the axial and tangential surface strain can be approximately calculated at any time, see Paper III.

![Fig. 14. Schematic of thermal fatigue test equipment. (Paper III)](image)

In a typical test, an induction unit (25 kW, 3 MHz) heats approximately 20 mm of the middle of the test rod, and continuous cooling is performed by circulating silicon oil (flow rate ≈ 2.5 l/min) of 60 °C through the specimen and also externally with argon or air. The external cooling with argon...
decreases also oxidation during testing. During thermal cycling, the surface temperature of the specimen is monitored by a pyrometer, but also measured by a K-type thermocouple with thin wires (Ø 0.13 mm), which is spot welded to the surface. The former method is used for temperature control during heating, and the latter to obtain recordings of the surface temperature during thermal cycling, which later is used in surface strain estimations.

Three temperature cycles were used to simulate various die casting temperature conditions, see Table 4. The temperature cycles included a steep ramp to the maximum temperature and subsequent cooling to the minimum temperature, with or without a short holding time (<0.1 s) at maximum temperature.

Table 4. Thermal cycles used in the thermal fatigue tests (Papers III, V, and VI).

<table>
<thead>
<tr>
<th>Max. temperature [°C]</th>
<th>Min. temperature [°C]</th>
<th>Heating time [s]</th>
<th>Total cycle time [s]</th>
<th>External cooling</th>
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<td>600</td>
<td>170</td>
<td>0.2</td>
<td>11.2</td>
<td>Argon or air</td>
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<tr>
<td>700</td>
<td>170</td>
<td>0.3-0.4</td>
<td>14.3-14.4</td>
<td>Argon or air</td>
</tr>
<tr>
<td>850</td>
<td>170</td>
<td>2.2-2.5</td>
<td>26.2-26.5</td>
<td>Argon or air</td>
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</tbody>
</table>

7.2 Surface temperature and surface strains during thermal cycling (Papers III-VI)

A representative surface temperature recording during thermal cycling to 700 °C is shown in Fig. 16. It is seen that the surface temperature increases rapidly during the short heating time, whereas the subsequent cooling proceeds more gradually.

![Temperature vs. Time](image)

Fig. 16. Representative surface temperature recording for the 700 °C tests in air at equilibrium temperature conditions. (Paper III)
Typical surface strain recordings during thermal cycling to 700 °C for a tool steel in quenched and tempered as well as surface engineered condition are shown in Fig. 17. It is seen that the surface strain recordings are composed of three phases. The surface strain increases with the surface temperature during the heating phase, followed by an almost constant or slightly increasing strain level during the initial steep cooling phase and, thereafter, it decreases with temperature. It is interesting to notice that recorded surface strain for surface engineered materials are comparable to those of the references. This indicates that the surface strain during thermal cycling predominantly is controlled by the deformation of the substrate material [Paper VI].

The appearance of the experimentally obtained surface strain recordings during thermal cycling, such as those of Fig. 17, were verified by numerical simulations of the thermal and mechanical conditions during heat cycling, see Paper IV.

Fig. 17. Typical surface strain recordings during thermal cycling to 700 °C in air at equilibrium temperature conditions. (a) QRO 90 at 510 HV30. (b) Surface engineered QRO 90 systems (tangential strain). (Paper VI)

The mechanical strain, which is necessary for any surface cracking, can be estimated based on the temperature and strain recordings, such as those of Figs. 16 and 17, see Paper III. A representative mechanical surface strain during thermal cycling to 700 °C is shown in Fig. 18. This type of information was used to estimate the stress in the surface layer, see Paper III, as well as to represent crack growth and thermal fatigue life using available strain-based models, see Paper V.
7.3 Surface cracking (Papers III and VI)

An example of thermal fatigue cracking after the 700 °C experiments is shown in Fig. 18. It is seen the reproduced crack pattern (Fig. 18a) resembles the type of surface cracking observed on tools in e.g. die casting applications, recall Fig. 7. In Papers III, V, and VI, the surface cracking was characterised with respect to crack length and density of cracks by measurements on polished cross-sections, such as that of Fig. 18b.

![Fig. 18. Example of thermal fatigue cracking after treatment with the 700 °C cycle in argon (borided specimens). (a) Typical crack network after 10 000 cycles. (b) Cross-section after 5 000 cycles. (Paper VI)](image-url)
7.4 Hardness after thermal fatigue (Papers III, V, and VI)

In Papers III, V, and VI, it was found that thermal fatigue above 700 °C cause considerable softening of the tool materials. Some examples of hardness vs. depth profiles after thermal cycling to 700 °C is given in Fig. 19. In this figure, it is seen that thermal cycling to 700 °C results in a hardness reduction of the superficial surface layer, and that the material deeper below is virtually unaffected.

In addition, an interesting observation from the hardness evaluations was that any initial ranking among various of tool materials is maintained throughout the tests, cp. Fig. 19 and see Papers III, V, and VI.

![Fig. 19. Example of hardness vs. depth profiles after 10 000 thermal cycles to 700 °C. (Papers III)](image-url)
8 Thermal fatigue cracking of hot work tool steels (Papers III and V)

Thermal fatigue cracking of the two hot work tool steel grades QRO 90 Supreme and Hotvar, hardened and tempered to various conditions, was evaluated using the thermal fatigue test mentioned above, see Papers III and V.

8.1 Stresses in the surface layer during thermal cycling (Paper III)

In Paper III, the stress in the surface layer during thermal cycling was estimated based on the mechanical strain loops and Hooke’s law in plane stress. It was found that the stress in the surface layer during thermal cycling to 700 °C varies as shown in Fig. 20. The thermally induced stresses cause accumulation of compressive plastic strains in the surface layer during the heating phase of the first cycle, cp. Fig. 20. The plastic yield of the tool material during the heating phase is, in turn, the prerequisite for development of tensile stresses in the tool surface layer during the cooling phase. Furthermore, the calculated tensile stresses imposed during cooling do not exceed the initial (true) yield stress of the tool material, cp. Fig. 20.

Even though the estimated tensile stresses never exceed the initial yield strength of the steel, it was found the numerous of thermal cracks are formed within the low-cycle fatigue range, see below. The explanation is that the presence of stress raising defects may cause the tensile stresses to exceed the tool steel yield stress during these conditions, and that the gradual softening of the tool steel that occurs during the thermal cycling degrades the initial yield strength values of the steel. Thus, the surface material locally will be exposed to cyclic stresses that cause accumulation of plastic strains after a certain number of thermal cycles.
8.2 Thermal fatigue crack length and crack density (Papers III, V and VI)

Thermal crack evaluations revealed that the crack length and crack density has a tendency to decrease with higher initial tool steel hardness [Papers III, V, and VI], as illustrated in Fig. 21. A higher hardness reduces the accumulation of plastic strains in the surface layer. The advantageous effect from higher hardness is maintained since any initial ranking in hardness among the steels is unaffected by the thermal cycling.

It was also found that an increase in the maximum cycle temperature promotes rapid the thermal crack nucleation and advance, as well as crack nucleation at larger number of sites [Papers III, V, and VI], cp. Fig. 21. The reason is that a higher maximum temperature promotes rapid softening, and increases the accumulation of plastic strain during each cycle.
Fig. 21. Maximum crack length (a) and crack density (b) after thermal cycling to 600 °C, 700 °C, and 850 °C, respectively, in argon. Each pile is the mean value of three or four specimens, except those for 100 cycles to 850 °C and 1 000 cycles to 700 °C, respectively. The error bars indicate the maximum and minimum recording. (Papers III and V)

8.3 Crack growth and thermal fatigue life (Paper V)

In Paper V, the cyclic thermal crack growth rate and strain intensity range was estimated using crack length observations and surface strain recordings. It was revealed that the crack growth rate and the strain intensity range during thermal cycling to 700 °C and 850 °C is of the order of
$10^{-9}$-$10^{-6}$ m/cycle and $10^{-5}$-$10^{-4}$ m$^{1/2}$, see Fig. 22. In addition, it was found that the crack growth rate increases roughly with one order of magnitude when increasing the maximum cycle temperature from 700 °C to 850 °C, revealing the strong temperature sensitivity mentioned above.

![Fig. 22. Crack growth rate $da/dN$ vs. strain intensity range $\Delta K$ for the tested tool steels during thermal fatigue to 700 °C and 850 °C. (Paper V)](image)

Furthermore, thermal fatigue life curves was constructed from the experimental findings, see Figs. 23 and 24. Note that $a_c$ is a critical maximum crack length that defines the number of cycles to thermal fatigue failure $N_f$. It is seen that the thermal fatigue life reduces roughly with one order of magnitude when increasing the maximum cycle temperature from 600 °C to 700 °C, and with an additional order of magnitude when increasing it from 700 °C to 850 °C. Again, revealing the strong temperature sensitivity.

This type of representation of the results from the thermal fatigue test, where the crack observations are coupled to the imposed load conditions, may aid in future comparisons of crack growth and thermal fatigue life of tool materials.
Fig. 23. Example of maximum cycle temperature vs. thermal fatigue life curve for the tested tool steels ($\alpha_c = 10$ µm). The minimum cycle temperature is 170 °C. The solid line is the power-law fit to the experimental data of QRO 90 (510 HV$_{30}$). Note that the scale is linear-logarithmic. (Paper V)

Fig. 24. Example of strain life curve for the tested tool steels ($\alpha_c = 10$ µm). The solid line is the power-law fit to the experimental data of QRO 90 at 510 HV$_{30}$. (Paper V)
9 Thermal fatigue cracking of surface engineered hot work tool steels (Paper VI)

An investigation on thermal fatigue cracking of a selection of surface engineered hot work tool steels, including common diffusion treatments (boriding, nitriding, Toyota diffusion (CrC)), PVD coatings (CrC, CrN, and TiAlN) and combinations of these, was conducted, see Paper VI. Untreated specimens of each tool steel were used as references.

9.1 Thermal cracking (Paper VI)

The thermal cracking characteristics of some of the investigated surface engineered tool steels is shown in Fig. 25. It was found that surface engineering generally have a tendency to reduce the resistance against thermal cracking as compared to the reference steels, as revealed by the longer crack lengths, see Fig. 25a and Paper VII. The explanation is that the mechanical properties of the tool material deteriorate during the engineering process. This is more thoroughly discussed in Paper VI.

Of the engineered layers evaluated, duplex-treatment improved the thermal cracking performance with respect to crack density, see Fig. 25b, and the single-layered CrN coating showed the best performance with respect to resistance to thermal crack penetration, see Paper VI. In fact, the resistance against thermal crack propagation of the single-layered CrN coating was almost comparable to that of the reference steel, see Paper VI.

An interesting observation was that all diffusion layers promote the thermal crack advance through their thickness. Consequently, the thicker the diffusion layer, the longer the maximum crack length [20,21], see Fig. 26.

Conclusively, the resistance against thermal crack penetration of surface engineered tool steels is determined by the mechanical properties of the substrate material.
Fig. 25. Maximum crack length (a) and crack density (b) after thermal cycling to 700 °C in argon. Each pile is the mean value of two to four specimens of each material, and the error bars indicate the maximum and minimum recording. (Paper VI)

Fig. 26. Maximum crack length vs. thickness of diffusion layer after thermal cycling. (a) 5 000 and 500 cycles to 700 °C and 850 °C, respectively. (b) 10 000 and 1 000 cycles to 700 °C and 850 °C, respectively. The solid lines are included to visualise any correlation. (Paper VI)
10 Corrosive wear in liquid aluminium (Paper VII)

Protective ceramic surface coatings on top of the tool material may be used in e.g. aluminium die casting applications to reduce the erosive and corrosive attack on the tool surface from the molten metal flow. In this thesis, the resistance against corrosion by molten aluminium of PVD CrN-coated tool steel specimens was investigated, see Paper VII.

10.1 Corrosion tests (Paper VII)

Cylindrical specimens, 90 mm in length, 13 mm in diameter, and with the tip rounded to a 6.5 mm radius, made of hardened and tempered Orvar Supreme were used as substrate material, see Paper VII. Six series of PVD coatings of CrN, representing six combinations of coating thickness and defect density, were applied on the substrate material.

The experimental set-up consisted of a resistively heated crucible with an aluminium alloy melt controlled to 710±10 °C. To maintain uniform temperature and concentration during testing, the molten aluminium was continuously stirred by a ceramic plate rotating at 80 rev./min. During each test, three specimens were submerged 1 h in the liquid aluminium.

10.2 Corrosion damage (Paper VII)

It was found that the corrosive attack is initiated through pinhole defects (detects through the coating), which act as small channels for the liquid aluminium through the coating down to the substrate, where it attacks the substrate material [Paper VII], see Fig. 27.

In an initial stage of development, the corrosive damage was observed in the form of localised circular bulges, see Fig. 27a. The bulge formation results from the corrosive attack on the substrate material below the defect, which cause a volume expansion of the substrate material due to formation of intermetallic compounds [31]. Continuous volume expansion of the
material below the defect results in local coating detachment when the deflection of the coating is excessively large. This results in a macroscopic corrosion damage observed in the form of circular corrosion pits, as well as pits that have grown together and formed larger cavities [31], see Fig. 28.

Finally, it was found that the resistance to corrosive wear, as the fraction of corroded area occupied by the pits, is improved by increasing the coating thickness, see Fig. 29. This is supported by the observations that the density of defects through the coating (pinhole defects) decrease with increasing coating thickness, see Paper VII.

![Fig. 27](image1.png)

**Fig. 27.** Typical bulge formation during the corrosion test after removal of solidified aluminium (SEM). (a) Top view. (b) Fractured cross-section. Note the pinhole defects in the centre of the bulges. (Paper VII)

![Fig. 28](image2.png)

**Fig. 28.** Macroscopic corrosive damage of the coated material identified after removal of solidified aluminium (SEM). (a) Overview revealing the type of bulge formation seen in Fig. 27, local corrosion pits, and larger corroded areas. (b) Local coating island without any corrosive damage (indicated by the arrow). (Paper VII)
Fig. 29. Fraction of corroded area vs. the coating thickness. (Paper VII)
11 Conclusions

The following major conclusions can be drawn from this doctoral thesis.

- A method to record the temperature fluctuations in a cavity insert during actual brass die casting was developed, and details of the thermal cyclic conditions were obtained.

- In actual die casting, the dominant tool failure mechanism is thermal fatigue cracking. The thermal crack formation is associated to accumulation of the local plastic strain in the surface that occurs during each casting cycle, typical of a low-cycle fatigue process. Initial crack growth is facilitated by oxidation of the crack surfaces, and proceeded growth is facilitated by this oxidation in combination with crack filling of cast material, and by softening of the tool material. In addition, local enrichment of Pb at the crack front from the cast alloy melt was observed to promote the crack growth in die casting of brass.

- A thermal fatigue test method to experimentally reproduce the characteristic type of surface cracking observed on die casting tools was developed. In addition, the test configuration enables recording of the surface strain during thermal cycling, which is impossible during actual die casting, and proved to give information of the strains and stresses behind the thermal fatigue failure.

- Thermal fatigue tests of tool steels revealed that low-cycle fatigue occurs, even though the calculated tensile stress after cooling does not exceed the yield stress that the material exhibits prior to testing. The reason is the gradual softening of the tool steel during the thermal cycling, and the presence of stress raising defects.

- Furthermore, the resistance against thermal cracking improves with initial tool steel hardness, and any initial ranking in hardness among the steels is unaffected by the thermal cycling.
• Thermal fatigue tests of surface engineered tool steels showed that surface engineering generally reduce the resistance against thermal cracking as compared to untreated references. The explanation is that the mechanical properties of the tool material deteriorate during the engineering process.

• Additionally, the resistance against thermal crack propagation of surface engineered tool steels is determined by the mechanical properties of the substrate material.

• Finally, the mechanisms of initiation and progress of liquid aluminium corrosion of CrN-coated tool steels have been explained. It occurs as follows: Initially, through-the-coating defects act as channels and cause the liquid aluminium to locally attack the steel. The subsequently formed corrosion pits coalesce and the corrosive attack aggravates.

• The liquid aluminium corrosion resistance of CrN coatings is improved by increasing their thickness.
12 References

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