Austempered High Silicon Steel

Investigation of wear resistance in a carbide free microstructure

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Abstract

The production of wood-pellets involves fractioning of the raw material, e.g. the sawdust and cutter shavings, into a homogeneous particle size before pressing through a ring die pelletizer. The fractioning takes place in a hammer mill, where wear on the cutting knives is considered a problem. The main material property requirements of the knives are high wear resistance and toughness. This research project examines the suitability of austempered high silicon steel to achieve these desirable properties. One specific steel, 55Si7, has been austempered to obtain a carbide free ausferritic microstructure and the mechanical properties, microstructure and wear resistance of this steel have been characterized. Microscopy and x-ray diffraction were used for microstructural characterization and examination of the stress-induced transformation of retained austenite into martensite. The mechanical properties were characterized via tensile testing, Charpy V-notch impact testing and hardness measurements; and the wear resistance has been investigated in a laboratory environment under non-lubricated sliding wear conditions. The wear resistance of austempered high silicon steel has been compared to other steels heat treated under different conditions to obtain different microstructures. In addition, a field test took place in a hammer mill for a direct comparison between the wear on four austempered high silicon knives and commercial knives made of case hardened low carbon steel. The austempered high silicon steel proved to have a uniform and high hardness combined with good impact toughness, plastic deformation before fracture, and high wear resistance. These desirable properties were achieved by austempering at low temperatures, resulting in a very fine microstructure and absence of carbides due to the presence of silicon.
Preface

The research presented in this final thesis reports on the results of work carried out at the Division of Engineering Materials at Luleå University of Technology (LTU) during the time period from January 2006 to June 2006. This thesis is, to the best of my knowledge, the result of original research except where reference is made to previous work, and constitutes the conclusive part of my degree in Mechanical Engineering at LTU. This has been carried out under the supervision of Esa Vuorinen at LTU.
Acknowledgements

First I would like to thank my supervisor Esa Vuorinen for his support and guidance throughout this project. I am grateful to Erik Navara, Per Rubin and Magnus Odén for their support and continual encouragement. I would like to thank Peter Hedström for helping out with interpreting XRD data, Johnny Grahn for the assistance with the SEM and Jens Carlevi for the assistance with wear testing. Finally I would like to express my sincere thanks to my family for always supporting me.
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1 Introduction

In the never ending search for new and improved materials, austempered high silicon steel stands out with its extraordinary combination of strength, hardness and toughness, created without the use of expensive alloying elements or mechanical processing. As a contribution to the on-going work of discovering applications and gaining an increased understanding of austempered high silicon steel, this research addresses the problem of abrasive wear occurring in sawdust cutting. Specifically, this research takes a direct comparison of the wear in current commercial knives against the wear on recently developed knives of austempered high silicon steel. Furthermore, the characterisation of the wear and mechanical properties of one specific steel, 55Si7, which has been heat-treated in different ways to obtain different microstructures, has also been studied.

1.1 Objective

The main objective of this research is to raise the public awareness of the possibilities of using austempered high silicon steels in applications where wear is a problem. In doing so, this research has focused on the characterisation of the mechanical-, microstructural- and wear-properties of high silicon steel, austempered to obtain a carbide free ausferritic microstructure.

1.2 Methods

The mechanical properties were characterized by tensile testing, Charpy V-notch impact testing and hardness measurement. Microscopy and x-ray diffraction were used to characterize the microstructure and study the stress induced transformation of retained austenite into martensite. The wear resistance was examined in laboratory environment where three austempered and thirteen reference specimens of various composition and heat-treatments were used. In addition to that a field test at Luleå Energi took place where four knives of austempered high silicon steel were installed in a hammer mill for a direct comparison with commercial knives made of casehardened low carbon steel.

1.3 Limitations

The investigation of mechanical and micro-structural properties has been limited to one specific type of steel with the chemical composition of Fe-0.56C-1.90Si-0.84Mn-0.19Cr-0.11Ni. Austenitization and austempering temperatures have been adopted from previous research and no further work has been done in this research to optimise the time and temperatures used.

1.4 Background

The production of wood pellets includes several processing steps in which the wear on the tool components is extensive. The knives used for cutting sawmill dust and the ring die used for pelleting the fragmented saw dust, are two examples where this extensive wear occurs. The following research has been limited to deal with the wear on the cutting knives.
Pellets have, in the last decade, become an economically interesting alternative to the use of electricity, fuel oil and firewood for heating purposes. The annual production of pellets in Sweden today is about one million tonnes, and is expected to increase in the coming years. Wood pellets, short cylindrical pieces of wood designed for heating, are manufactured by compressing dry, fine particles of raw material, such as sawmill dust and cutter shavings (Figure 1). The raw material for pellet production comes almost entirely from forest industry by-products, such as cutter shavings, sawdust and bark. Until recently, these by-products were mainly used to make chipboards; however, production of chipboard decreased dramatically in the early nineties which liberated the raw material for the alternative fuel market [1].

![Figure 1 Wood pellets, made of compressed sawmill dust](image)

The hammer mill, with a diameter of approximately 1 metre, contains 72 knives, with dimensions of 160 x 60 x 6 mm. These knives are distributed over four shafts that rotate at a speed of 1400 rpm. Each of the knives four corners are used before being replaced (Figure 2). Once the first corner has been worn out, the direction of the rotation is changed. The third and fourth corners can also be used after the knives are demounted and turned.

![Figure 2 Knives for sawdust cutting, upper knife shows typical wear before replacement.](image)

The wear on the knives is abrasive in nature, caused by sand contamination in the raw material which occurs in the handling and storage procedures. Occasionally, larger
objects such as screws, bolts and small rocks are fed into the hammer mill with the sawdust. Given the high rotational speed, the impact force caused by these larger objects could cause the knives to rupture, resulting in severe and costly damages to the machine. A high toughness is therefore of great importance in order to avoid this rupturing.

High wear resistance and toughness can be achieved by different solutions such as cladding, case hardening, or the use of steel with a high uniform hardness. The knives used for the fractioning of sawdust in the hammer mill investigated in this research are made of case hardened carbon steel, which is a common way of obtaining a hard, wear resistant surface at a reasonable cost. The hardness of the surface layer is around 900 Vickers, which gradually decreases down to 300 Vickers, at about a depth of 1 mm into the material.

Austempered high silicon steel comes up as an interesting alternative to evaluate, as both the costs of the raw material, and heat treatment are comparable to the steel currently in use. The austempered steel is advantageous in that a uniform hardness of 500 Vickers can be achieved without compromising the impact toughness. Positive results from previous research [2, 3] of austempering of the 55Si7 steel, demonstrated that sufficient toughness and hardness could be obtained to satisfy the requirements for the knives. As a result of this, a decision was taken to manufacture five knives to be installed and tested in industrial production. The wear progress and outcome of this test has been evaluated in this research.
2 Austempered High Silicon Steel

Bainite is a non-equilibrium transformation product of austenite which is produced by cooling at rates such that the diffusion-controlled transformations of pearlite are not possible, yet the cooling is sufficiently slow to avoid the diffusion-less transformation into athermal martensite. This can be achieved by either continuous cooling or by austempering. Austempering involves austenitizing the steel followed by quenching to a temperature below the pearlite nose, but above the martensite start temperature, $M_s$, where isothermal transformation takes place until all of the austenite transforms to bainite (Figure 3).

![Figure 3 TTT and CCT diagram, calculated for a 55Si7-steel (4).](image)

The objective of austempering is the formation of bainite, which is similar in form and has a comparable hardness to martensite tempered at the same temperature; however it has a higher toughness and lower residual stresses than the corresponding martensite. In plain carbon steel, bainite is formed in two stages. In the first stage, ferrite laths nucleate and grow in carbon enriched retained austenite, followed by a second stage where carbides precipitate after the austenite has become saturated with carbon.

Stage I  
\[ \gamma \rightarrow \alpha + \gamma_{hc} \]

Stage II  
\[ \gamma_{hc} \rightarrow \alpha + \text{carbide} \]

The carbides can precipitate either within the ferrite or at the ferrite/austenite boundaries. The terms upper and lower bainite are used to distinguish between these two under-types. Upper bainite is formed at higher temperatures where the diffusion rates are high enough for carbon to diffuse to the lath boundaries. Lower bainite forms at lower temperatures where the carbon is trapped within the growing ferrite lath resulting in carbide precipitation within the ferrite [5].

Silicon is virtually insoluble in iron carbide. Hence, it is necessary for silicon to diffuse out of the forming carbide particles, which is a process slower by many orders of magnitude than the diffusion of interstitially dissolved carbon atoms. This difference in diffusion coefficients between carbon and silicon in iron is the reason for silicon’s retarding influence on the second stage of the bainitic transformation. The difference
increases with decreasing temperature, as the activation energy of substitutional diffusion far exceeds that of interstitial diffusion. Since the second stage in the bainite formation is suppressed in high silicon steel, the term bainite does not correctly describe the obtained structure after austempering in the bainitic temperature range [6]. The term bainite implies presence of carbides and thus, a structure that consists of a carbide free blend of acicular ferrite and austenite is better described by the term ausferrite. The principle difference in microstructure between bainite and ausferrite is schematically illustrated in figure 4.

![Figure 4](image)

**Figure 4** Schematic illustration of the microstructural differences obtained by austempering of plain carbon steel and high silicon steel. When the second stage reaction for bainite is completed, the upper and lower bainite consists of only ferrite and carbides. In high silicon steel, no carbides form and the microstructure is then termed ausferritic.

### 2.1 Formation of bainite

The formation of bainite is still not fully understood, resulting in incompatible theories and confusion of languages. The nucleation and growth of bainitic ferrite has been subject to controversy for decades and different schools of thought have developed. The two opposing viewpoints that dominate the debate are whether the mechanism is diffusive or displacive in nature [6]. The following is intended to give insight into the mechanism behind bainite formation and the above mentioned debates concerning this formation. The discussion concerning nucleation and growth of bainite also applies to the formation of ausferrite since the second stage of the process where carbides form is not dealt with.

A phase transformation can only occur when a lower state of energy is achieved by the redistribution of atoms. Equilibrium condition can be said to apply when the Gibbs free energy is minimized. A bainitic microstructure is far from equilibrium. For example, the free energy change accompanying the formation of bainite in a Fe-0.1C wt% alloy at 540°C is approximately -540J/mol, whereas the free energy change accompanying the formation of a (para-) equilibrium mixture of allotriomorphic ferrite and austenite at the same temperature is approximately -1050 J/mol. Consequently, the excess energy of
bainite could in this case be said to be approximately 470J/mol and equilibrium conditions are therefore not fulfilled [5].

Bhadeshia points out that it is impossible to say when a system has reached equilibrium in its lowest minimum of free energy state [5]. He therefore claims it is more realistic to refer to the state of metastable equilibrium, which represents a local minimum in free energy but does not exclude the existence of further lower minima. Para-equilibrium in the context of bainite is defined by Bhadeshia as a constrained equilibrium which arises in substitutionally alloyed steels, in which the ratio of iron to substitutional solute atoms remains the same everywhere, but subject to this constraint, the carbon atoms achieve equality of chemical potential in the parent and product phases. In diffusion controlled transformation products, such as pearlite, it can be assumed that a local equilibrium exists at the interface between two phases in contact even though the phases may not be homogeneous. Bhadeshia also clarifies that between local- and para-equilibrium, there could exist an infinite number of states in which the substitutional elements partially partition between the phases, which explains the persistent controversy that exists about the exact nature of the mechanism of nucleation and growth of bainite.

During growth of bainite it is clear that the austenite becomes carbon enriched and that the carbon is rejected from the ferrite laths. What is not clear is when the carbon is rejected. In principle there are three possibilities of how the carbon can behave during bainite formation.

1. The carbon may partition during the growth so that the ferrite may never contain any excess carbon. (Diffusive)
2. The growth may proceed with carbon being trapped within the ferrite lath by the advancing interface, resulting in a supersaturated ferrite where carbon is rejected into the retained austenite when growth of the ferrite lath is terminated. (Displacive)
3. Some carbon may partly diffuse during growth with the remainder being trapped and the ferrite grows partially supersaturated.

According to the diffusive mechanism, some short-range diffusion of the lattice atoms is required. Originally, this was proposed to occur at well developed ledges in the $\alpha/\gamma$ interface. The modelling of the interface structure has now grown more sophisticated and the term diffusive is a general term that includes any structural feature where lattice atoms can cross the interface by random jumps [7], in contrast to the displacive mechanism where no random jumps of lattice atoms can occur.

According to the displacive mechanism, the $\alpha/\gamma$ interface is of a glissile, martensitic type which can move without any random jumps of the lattice atoms. In principle, the movement may be, slow, controlled by long-range diffusion of carbon, or, very rapid as in a martensitic transformation [7].

By measuring the local compositions before and after the transformation it has been established that martensitic transformation is diffusionless. The chemical composition of the martensite is identical to the parent austenite [8]. Bainite forms at somewhat higher temperatures where the carbon can escape out of the ferrite lath within a fraction of a second and a direct measurement of its original composition can therefore not be made. It is therefore much more difficult to determine the precise role of carbon during the growth of bainitic ferrite than in martensite.
In the early research of bainite formation, Zener [9] proposed that bainite forms as a succession of plates, growing very quickly to its full size and without giving carbon time to partition. After the formation of one plate, the carbon then diffuses into the austenite and the next plate of bainite would thus have to form from a γ phase with higher carbon content, which should be thermodynamically more difficult. Zener proposed the final state where no more bainite forms, should occur when the carbon content of the austenite has reached the $T_0$ line, see figure 5. This theory is among others supported by Hillert and Ågren noting that the possibility of reconstructive (diffusive) transformation can occur at temperatures where the chemical diffusivity is low, and therefore should not be ruled out [10].

![Figure 5 Illustration of carbon enrichment of austenite during isothermal transformation, where $\overline{x}$ is the mean carbon concentration of the steel [5].](image)

The variation in composition in the vicinity of the transformation interface for different growth mechanisms depends on the solute atoms ability to diffuse, as illustrated in Figure 6 [5]. The diffusion rates of carbon and substitutional solute atoms (X) at the austempering temperatures for bainite formation may differ by several orders of magnitude, only allowing for diffusion of carbon. In an alloy in which there is atomic mobility within the parent phase, the reconstruction (displacement) provides an opportunity for the solvent and solute atoms to redistribute between the two phases. Partitioning local equilibrium (P-LE) growth of ferrite from austenite is controlled by diffusion of X-atoms. It is a relatively slow growth mechanism with a concentration gradient of X-atoms ahead of the interface. In the case of a rapid growth of ferrite in austenite, negligible partitioning local equilibrium (NP-LE) applies, where the partitioning of X-atom will be extremely small with a near uniform concentration of X, except for a small spike in the austenite adjacent to the interface. As the ratio of the C/X diffusion rates increases, the width of this spike decreases and the concept of paraequilibrium applies. Further increase in the rate of formation will lead to partial supersaturation of carbon in the ferrite. Martensite is formed without diffusion of either carbon or X atoms.
It is well established that no long range redistribution of substitutional (X) alloying elements occurs during bainite growth, so that the X/Fe atom ratio is the same in the parent and product lattices [11]. Partitioning local equilibrium can therefore be ruled out. It logically follows to question whether short range X atom concentration fluctuations exist or not in the vicinity of the transformation interface. In situ-observations of lattice parameter fluctuation in austenite during formation of ferrite has shown that the bainitic ferrite forms in carbon-depleted zones in the parent austenite, with carbon trapped in the ferrite in the early stages of formation. The estimated diffusion distance for substitutional atoms was found to be less than the inter-atomic distance and the lattice parameter fluctuation in austenite could therefore not be
attributed to diffusion of substitutional atoms [12]. This implies that bainite grows with partial supersaturation of carbon in ferrite and without diffusion of lattice atoms.

2.2 The $T_0$ concept

With the assumption that bainitic ferrite grows without diffusion, it is possible to use the $T_0$-concept to theoretically determine when the formation of ferrite laths no longer can proceed. By predicting the maximum amount of ferrite that can be achieved at a given temperature, an estimation of the time needed to obtain a fully developed ausferritic microstructure is facilitated. The maximum amount of carbon that can be stored in the retained austenite is schematically explained by the $T_0$-theory. Growth by a diffusionless mechanism has to occur at a temperature below $T_0$, when the free energy of the ferrite falls below the free energy of austenite (Figure 7). Therefore ferrite is allowed to form without diffusion (displacive), as long as the carbon concentration in the austenite is below the $T_0$-line, whereas for concentrations above $T_0$ diffusionless transformations cannot occur. It should be noted that diffusive and displacive in the context of bainite refers to the mobility of iron and not the interstitially soluted carbon. There is therefore an apparent paradox in that a displacive transformation in which there is segregation of carbon during growth has to be described as non-diffusional even though the growth rate might be controlled by the diffusion rate of carbon!

When a sheave of ferrite is formed, the excess carbon is rejected into the surrounding austenite, where the carbon concentration consequently increases. The next sheaf thus grows from carbon enriched austenite. This process can only continue until the carbon concentration in the austenite equals that of the $T_0$-curve, and is therefore said to be an incomplete reaction since it is terminated before it reaches the equilibrium composition for the austenite.

If, on the other hand, growth of ferrite with a carbon concentration equal to the $T_0$-composition would be allowed, the process would proceed until the austenite composition equals that at the $A_{c3}$-line [13]. The ferrite can only hold a limited amount of carbon, given by the $A_{c1}$-line; however, if carbides are formed within the ferrite laths, the mean carbon concentration in the ferrite can be held constant at the $T_0$-concentration, leading to a complete reaction. Nonetheless, the formation of carbide is prevented if more than approximately 1.5 wt% silicon is added to the steel, keeping it carbide free and hence the transformation can be expected to stop as the carbon level in austenite reaches $T_0$. 
Figure 7 Schematic illustration of the $T_0$ and $T'_0$-curve and the Gibbs free energy curves for ferrite and austenite at a temperature below the $A_1$-temperature. $A_{e1}$ and $A_{e3}$ are extrapolated curves of the $A_3$ and $\alpha$-ferrite curves in the Fe-C phase diagram [14].

It has been found that in some cases, the carbon concentration exceeds that predicted by the $T_0$-concentration, even though no carbides have been found, which, from a thermodynamical point of view, is impossible. This has been a source of legitimate criticism of the theory that bainite transformation is diffusionless. One explanation of this incompatibility is that the mechanism in which the bainite grows is without any diffusion, but excess carbon is rejected into the austenite soon after transformation [15]. The stored energy in bainitic ferrite is accounted for by raising its free energy curve by an amount equal to the strain energy due to transformation, giving the $T'_0$-curve.

Further analysis concerning the different viewpoint of the formation of bainite is beyond the scope of the present work and will be left for the curious truth-seekers to discover elsewhere.
3 Experimental procedures

A spring steel, EN 1.5026 (from Fundia, Smedjebäken), has been used for all of the austempering experiments. The steel was delivered as hot-rolled flat bars, 12 mm thick and 65 mm wide. The chemical composition was determined to 0.56C-1.90Si-0.84Mn-0.19Cr-0.11Ni (wt%). Austenitization was carried out for fifteen minutes at 900°C and austempering of the steel in a salt bath oven for two hours has been used in all experiments, unless otherwise noted.

The mechanical properties were investigated by tensile-, hardness and impact toughness testing. All tests were carried out at Luleå University of Technology. A Dartec Universal Straining Frame, fitted with a 250 kN load cell pulling with a low strain rate of 0.1 mm/s, was used for tensile testing, where the elongation was recorded via an electronic data acquisition system coupled with an extensometer with a gauge length of 50 mm. Three tensile test specimens having a gauge length of 55 mm and a gauge diameter of 5 mm were turned in a lathe. The impact toughness was measured in a Charpy V-notch testing machine with a capacity of 160 J. The V-notched impact test specimens were machined to dimensions of 10 x 10 x 60 mm. A total of six specimens were impact tested. A Matsuzawa MXT-CX microhardness tester was used for all hardness measurements.

Optical and scanning electron microscopy were used to examine etched microstructures. Specimens were ground with diamond paste and polished with a 0.25 μm silica solution before etching with a 2% nital. The SEM equipment used was a Jeol JSM 6460 LV fitted with Oxford Inca EDZ software.

Quantitative x-ray diffraction analysis was used to determine the volume fraction of retained austenite after austempering. The samples were step scanned in a Philips MRD, operating at 40 kV voltage 45 mA current, using Cu Kα radiation. The 2θ scan angles ranged from 20° to 100°. The data was collected in a computer file and analyzed using Origin 7.5 SR1 software. The carbon concentration in austenite and ferrite was estimated using obtained lattice parameters from the x-ray diffraction patterns.

The tribological behaviour of the austempered silicon steel and a selection of low alloy and boron steels were examined in laboratory conditions using a cameron plint reciprocating wear testing machine. A total of 114 wear tests were performed on 16 different samples. All tests were performed in unlubricated conditions at room temperature. The sliding stroke was 13 mm and was maintained at a rotating speed 44 rpm for a duration of 60 minutes, resulting in a total sliding distance of 68.6 m. Two different loads were applied, 20 and 300 N in order to determine whether the wear behaviour was altered between different loads.

In order to replicate wear, steel roller-bearing balls, SKF RB6/G20W, with a hardness of 860 HV0.3 were used as wearing parts. A new ball was used for every test. The wear samples were ground and polished and prior to testing they were cleaned in an ultrasonic cleaner over two steps (using n-heptane and acetone) to ensure no contamination on the surfaces would influence the wear process. The amount of wear was quantified by measuring the weight lost of both the wear specimens and the bearing balls, using a fine scale. Qualitative examination of the wear scar was made with a scanning electron microscopy. To measure the depth and geometry of the wear scars, a WYKO NT 1100 3-D topography profiler was used.
A field test of cutting knives in production equipment took place at Luleå Energi AB in Luleå. Four knives were manufactured and heat-treated to obtain the desired ausferritic microstructure required for this study. The knives were mounted in a hammer mill, along with the original knives in order to obtain a direct comparison of the wear that occurred during the cutting process of the sawmill dust.

The hammer mill has a diameter of approximately 1 metre. Within the mill there are 72 knives, with the dimensions of 160 x 60 x 6 mm, which are distributed over four shafts that rotate at a speed of 1400 rpm. Each of the four corners of the knives were used before the knives were considered to be worn out, first by changing the direction of rotation, then by demounting and turning the knife upside down. The hammer mill was regularly stopped for visual inspections of the wear progress. Quantitative determination of the wear was made using image analysis and weight loss measurements were made when the knives were demounted for turning and after all four corners were worn out.
4 Results and discussion

4.1 Mechanical properties

The mechanical properties were obtained by tensile- and Charpy V-notch testing. Austempering at 340°C proved to provide a significant increase in strength and hardness compared to the as rolled steel, and maintained the ability of plastic deformation (Table 1). Tensile strength exceeding 1500MPa; a hardness of close to 500 Vickers; and a Charpy V-notch impact toughness of 20 J cm\(^{-2}\) were obtained. Similar values have been reached by researchers using a 0.5C-1.5Si steel austempered at 350°C [16].

The desired mechanical properties can be attributed to the very fine microstructure obtained after austempering. The microstructure consists of fine laths of ferrite, separated by thin films of austenite. The ferrite laths were found to be 10 to 20 μm long and approximately 0.2 μm wide, as can be seen in Figure 8. The high toughness in ausferrite is produced when a high degree of transformation of austenite to ferrite is possible, leading to a fine microstructure of ferrite laths separated by thin films of stable austenite. With low transformations of austenite, a blocky morphology of less stable retained austenite is obtained which is detrimental to the toughness [17, 18].

Table 1 Effect of heat treatment by austempering 2 hours at 340°C. Before heat treatment, indicates the mechanical properties of the steel as delivered from steel plant after hot rolling.

<table>
<thead>
<tr>
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<th></th>
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<tbody>
<tr>
<td>(R_p^{0.2})</td>
<td>612 MPa</td>
<td>1250 MPa</td>
</tr>
<tr>
<td>(R_m)</td>
<td>1047 MPa</td>
<td>1530 MPa</td>
</tr>
<tr>
<td>(A_5)</td>
<td>13 %(^1)</td>
<td>12 %(^2)</td>
</tr>
<tr>
<td>(Z)</td>
<td>40 %</td>
<td>58.5 %</td>
</tr>
<tr>
<td>Hardness</td>
<td>338 HV(_{0.3})</td>
<td>485 HV(_{0.3})</td>
</tr>
<tr>
<td>Impact toughness (+20°C)</td>
<td>12 J</td>
<td>20.7 J</td>
</tr>
</tbody>
</table>

1 Fracture outside the measuring range
2 Only one valid measurement
Six specimens were used for the Charpy V-notch testing at room temperature. All specimens had been austempered at 340°C. Three of the specimens were tested without further treatment and the remaining three specimens were tested after being exposed to cryogenic heat-treatment by deep freezing in liquid nitrogen prior to the testing. There was no significant difference seen in the impact energy between the samples, indicating that the stability of the retained austenite is very high. Transformation of retained austenite into martensite would be expected to embrittle the specimen and thus lower the impact energy. With no significant decrease in impact energy it can be expected that none, or only small amounts of the retained austenite could have transformed into martensite after the cryogenic heat treatment.

The fracture surfaces show a combination of transgranular quasi-cleavage fracture in combination with equiaxed dimples from microvoid coalescence (Figure 9). Microvoid coalescence indicate ductile behaviour and occurs by formation of microvoids that form upon discontinuities in the steel. These voids grow during deformation and eventually join up together and with the crack tip, which leads to failure. Transgranular fractures are less ductile than those caused by microvoid coalescence. Transgranular fractures follow the lattice edge resulting in a fairly smooth looking fracture. No intergranular "mirror like" fractures, that would indicate brittle fracture, could be found. This corresponds well with the obtained results from impact testing and the ductile behaviour of the tensile test specimens before fracture. SEM investigation of specimens austempered at 320°C and impact tested at -40°C [2] demonstrated that the transgranular fracture surfaces had increased at the expense of microvoids, resulting slightly lower impact energy values. No brittle intergranular surfaces were detectable, meaning the ductile behaviour can be expected to be preserved at low temperatures, a result that would be of great interest for applications operating in outdoor winter climate.

Figure 8 Ausferritic microstructure, a very fine mixture of ferrite and austenite. SEM picture, taken of specimen austempered two hours at 340°C.
After austempering specimens of the 0.56C-1.9Si steel for one hour at 320°C and 500°C the hardness was found to be 500 and 295 Vickers, respectively [2]. The impact strength of the harder specimens proved to be consistently higher in the temperature interval spanning from -40 to + 60°C (Figure 10). This can be explained by the more blocky, and thus less stable, austenite expected to be received with austempering at higher temperatures. Blocky morphology exhibits a triangular shape in two dimensional cross sections while thin films of austenite are found between the subunits in sheaves of ferrite. The retained austenite content present after completed transformation decrease as the transformation temperature is decreased and thus permit more of the austenite to be in the film morphology. It has been suggested that the ratio of volume fraction between film/blocky morphology must exceed 0.9 for optimum toughness-strength combination [17]. The influence of the austempering temperature on the mechanical properties has been well documented [16, 19, 20, 21, 24]. By choosing a high silicon steel with a carbon content of 1 wt% and lowering the austempering temperature, hardness values exceeding 600 Vickers and ultimate tensile strengths of 2500 MPa have been reached, together with an ability to flow plastically before fracture [21].
4.2 X-ray diffraction

X-ray diffraction of crystalline materials produces a diffraction pattern of unique peaks that correspond to the present phases of the investigated material. The diffraction pattern is displayed as a graph of intensity as a function of diffraction angle \( \theta \), and can be used for qualitative and quantitative phase identification and analysis.

4.2.1 Carbon in retained austenite and ferrite

X-ray diffraction has been used to determine the amount of retained austenite and carbon in the ferrite. The volume fraction of retained austenite was calculated from the integrated intensities of the (110), (200), (211) and (220) ferrite peaks and (111), (200), (220) and (311) austenite peaks (Figure 11). Quantitative phase analysis is based on the premise that crystalline materials have unique diffraction patterns and that the intensity of the peaks in each particular pattern varies directly with its concentration. There are many factors that prevent the direct comparison of concentration through peak intensity. The basic cause preventing this is the different x-ray absorption properties of the substances in the sample [22]. By integrating the peaks, without taking into account for the absorption influence, the absolute values become of secondary interest, while the primary interest lies within the comparison between samples. Three tests were performed in order to see how the volume fraction of ferrite increases at lower austempering temperatures. The chosen temperatures were 340, 315 and 295 degrees Celsius.

Figure 10 Results from Charpy V-notch testing. Austempering at higher temperature resulted in lower impact strength. [2]
Integration of the peaks showed that the volume fraction of austenite indeed decreased as the temperature was lowered, see Table 1. This means that at lower temperatures, higher carbon content in the retained austenite is attained before the transformation is terminated, as suggested by the $T_0$ theory.

### Table 2 Volume fraction of retained austenite, determined from integrated intensities of (110), (200), (211) and (220) ferrite peaks and (111), (200), (220) and (311) austenite peaks after x-ray diffraction.

<table>
<thead>
<tr>
<th>Austempering temperature</th>
<th>340°C</th>
<th>315°C</th>
<th>295°C</th>
</tr>
</thead>
<tbody>
<tr>
<td>$V_\alpha$ (%)</td>
<td>80</td>
<td>86</td>
<td>92</td>
</tr>
<tr>
<td>$V_\gamma$ (%)</td>
<td>20</td>
<td>14</td>
<td>8</td>
</tr>
</tbody>
</table>

The carbon concentration in the retained austenite has been estimated with the use of the lattice parameters received from XRD. Equation 1 describes the relationship between the austenite lattice parameter, $a_{FCC}$, and the alloying elements at room temperature [23].

$$
a_{FCC} = 3.5780 + 0.033x_C + 0.00095x_{Mo} + 0.0002x_{Ni} + 0.0031x_{Mo} + 0.0006x_{Cr} + 0.0018x_\gamma + 0.0056x_{Al}$$  

\text{eq. 1}

In the above equation, $x_i$ corresponds to the weight percent of elements “i” in the austenite. By using the relationship between the atomic plane distance and the lattice parameter $a$ in cubic structures, equation 2, together with Bragg's law, equation 3, a relationship between two theta and the carbon content can be obtained. Using this relationship, a carbon content of 1.75 wt% best fit the four austenite peaks from XRD results.
Similarly, the carbon content in the bainitic ferrite can be calculated by using the relationship between the ferrite lattice parameter and the alloying elements, equation 4 [23]. The calculated carbon content in bainitic ferrite is then 0.15 wt%, which is in accordance with measured amounts of carbon in ferrite after austempering at similar temperatures [24].

\[
a_{\text{BCC}} = 2.8664 + \left(\frac{a_{Fe} - 0.279M_{C}}{a_{Fe} + 2.496M_{C}}\right)^{2} - a_{Fe}^{3} - 0.03M_{Si} + 0.006M_{Mo} + 0.07M_{Ni} + 0.31M_{Mo} + 0.05M_{Cr} + 0.096M_{Y}
\]

The sum of carbon in ferrite and the stabilized austenite equals the total amount of carbon in the steel according to equation 5 and therefore the volume fraction of austenite and ferrite can be estimated by using equations 6 and 7.

\[
\bar{x} = x_{\gamma}V_{\gamma} + x_{\alpha}V_{\alpha}
\]

\[
V_{\alpha} = \frac{\bar{x} - x_{\gamma}}{x_{\alpha} - x_{\gamma}}
\]

\[
V_{\gamma} = \frac{x_{\alpha} - \bar{x}}{x_{\alpha} - x_{\gamma}}
\]

\[
\bar{x} = \text{mean carbon wt}\% \\
x_{\gamma} = \text{carbon wt}\% \text{ in retained austenite} \\
x_{\alpha} = \text{carbon wt}\% \text{ in ferrite} \\
V_{\alpha} = \text{Volume fraction ferrite}
\]

Solving these equations for 1.75 wt% C in austenite and 0.15 wt% C in ferrite, gives that the volume fraction of ferrite is 74%. When compared with result from the integrated peak area, the difference is roughly 6%. However, the relationship between intensity and concentration is not linear because the diffracted intensity depends markedly on the absorption coefficient of the phases, which itself varies with the concentration of the phase. In order to accurately determine the 2θ angle and the relative intensity of the diffracted x-ray beams, it would be necessary to perform a comparison test with a sample of known composition. Preferred orientation of the grains from sample preparation, instrumental misalignment and specimen displacement are sources of error that could influence the XRD pattern, as can be seen in Figure 12. A shift of 2θ by 0.05 degrees corresponds to roughly 0.1 wt% C in austenite and ferrite when using the above equations and a considerable margin of error in the calculation can therefore be expected. The difficulty to accurately determine the retained austenite content could explain the large variations, from 10 to 40%, reported for similar steels [16, 25].
4.2.2 Influence of carbides and martensite

If carbides have formed within the ferrite sheaves or some of the retained austenite transforms into martensite, it will influence the volume fractions of austenite and ferrite. Equation 8, where $x_c$ and $V_c$ denotes the carbide carbon content and volume fraction respectively, can be used to calculate the volume fraction of retained austenite if carbides are present.

$$V_\gamma = \frac{\bar{x} - x_\alpha + V_c (x_\alpha - x_c)}{x_\gamma - x_\alpha}$$

No evidence of carbides has been found using SEM or XRD and it can be assumed that the silicon has prevented carbide formation. However, it can not be excluded that carbides have formed in such small amounts that it would difficult to detect using SEM and XRD. Even small amounts of carbides would influence the amount of retained austenite. Figure 13 illustrates how the volume fraction of retained austenite varies between roughly 10 to 30 % with only small alterations in austenite and ferrite carbon concentration and a maximum of 2 % carbides.
Figure 13 Illustration of how the amount of retained austenite depends on ferrite carbon content, carbides and martensite. The curves are plotted using equations 7 and 8 and show how the volume fraction of retained austenite varies between roughly 10 to 30% with only small alterations in austenite and ferrite carbon concentration and a maximum of 2% carbides.

The volume fraction of austenite would also decrease if martensite is present. Martensite forms with the exact same composition as the retained austenite and therefore 1% of martensite would further decrease the austenite content with one percentage unit. For 1.75 wt% C, the a and c-lattice parameters of BCT martensite are 2.84 and 3.07 Å respectively, if calculated using equations 9, 10 [27]. The tetragonality of the ordered martensite, measured by the ratio between the lattice parameters, c/a, increases with the carbon content, and for 1.75 wt% the c/a ratio becomes 1.079.

\[ a_m = 2.867 - 0.013 \quad \text{eq. 9} \]

\[ c_m = 2.867 + 0.116 \quad \text{eq. 10} \]

Finding martensite with XRD is a difficult task since its peaks are expected to be found at the same, or very close to the diffraction angle of ferrite and austenite (Table 3). Only the diffraction peaks from the (002) and (112) planes are distinctly separated from the austenite and ferrite peaks.
Table 3 Lattice parameters, atomic plane distance and 2θ-position calculated for 0.15 wt% carbon in ferrite and 1.75 wt% carbon in austenite and martensite. The only position where a separate peak for martensite is to be found is for the (002) and (112) planes.

<table>
<thead>
<tr>
<th>(h,k,l)</th>
<th>d (Å)</th>
<th>2θ</th>
</tr>
</thead>
<tbody>
<tr>
<td>Ferrite</td>
<td></td>
<td></td>
</tr>
<tr>
<td>(1,1,0)</td>
<td>2.028</td>
<td>44.645</td>
</tr>
<tr>
<td>(2,0,0)</td>
<td>1.434</td>
<td>64.979</td>
</tr>
<tr>
<td>(2,1,1)</td>
<td>1.171</td>
<td>82.275</td>
</tr>
<tr>
<td>(2,2,0)</td>
<td>1.014</td>
<td>98.865</td>
</tr>
<tr>
<td>Austenite</td>
<td></td>
<td></td>
</tr>
<tr>
<td>(1,1,1)</td>
<td>2.100</td>
<td>43.081</td>
</tr>
<tr>
<td>(2,0,0)</td>
<td>1.818</td>
<td>50.170</td>
</tr>
<tr>
<td>(2,2,0)</td>
<td>1.286</td>
<td>73.679</td>
</tr>
<tr>
<td>(3,1,1)</td>
<td>1.096</td>
<td>89.346</td>
</tr>
<tr>
<td>(2,2,2)</td>
<td>1.050</td>
<td>94.500</td>
</tr>
<tr>
<td>Martensite</td>
<td></td>
<td></td>
</tr>
<tr>
<td>(1,0,1)</td>
<td>2.081</td>
<td>43.367</td>
</tr>
<tr>
<td>(1,1,0)</td>
<td>2.006</td>
<td>45.077</td>
</tr>
<tr>
<td>(0,0,2)</td>
<td>1.532</td>
<td>60.293</td>
</tr>
<tr>
<td>(2,0,0)</td>
<td>1.419</td>
<td>65.650</td>
</tr>
<tr>
<td>(1,1,2)</td>
<td>1.217</td>
<td>78.363</td>
</tr>
<tr>
<td>(2,1,1)</td>
<td>1.173</td>
<td>81.902</td>
</tr>
<tr>
<td>(2,0,2)</td>
<td>1.041</td>
<td>95.286</td>
</tr>
</tbody>
</table>

However no indication of martensite could be detected after austempering. The martensite starting temperature, \( M_s \), depends on the amount of carbon in the austenite. With increased concentrations of carbon, the temperature is lowered when the transformation starts, which is approximately in accordance to the following relationship [28]

\[
M_s(K) = 764.2 - 302.6w%C - 30.6w%Mn - 16.6w%Ni - 8.9w%Cr + 2.4w%Mo - 11.3w%Cu + 8.58w%Co + 7.4w%W - 14.5w%Si
\]

For a chemical composition of 0.56C-1.90Si-0.84Mn-0.19Cr-0.11Ni, the calculated \( M_s \) temperature becomes 265°C. As the carbon concentration increases in the retained austenite during the bainite transformation, the \( M_s \) temperature will decrease and at carbon concentrations of about 1.3 wt%, the \( M_s \) will be below room temperature. The calculated \( M_s \)-temperature for 1.75 wt% carbon in retained austenite is -95°C. A specimen that had been austempered at 340°C was deep frozen in liquid nitrogen (-196°C) in an attempt to induce martensite transformation. After deep freezing the specimen was investigated with X-ray diffraction. This showed that the austenite peaks was virtually unchanged, but small disturbance at the foot of the austenite and ferrite peaks indicated the presence of small amounts of martensite. If such small amounts are transformed after deep freezing in liquid nitrogen it can be stated that there is no martensite present after austempering at 340°C and cooling to room temperature.

The x-ray diffraction result thus tells that austempering at 340°C for 2 hours has produced an ausferritic microstructure free of embrittling carbides and martensite.
4.3 Wear test

The wear resistance was tested in laboratory environment where three austempered specimens and thirteen reference specimens of various composition and heat treatments were used. In addition to that a field test at Luleå Energi took place where four knives of austempered high silicon steel were installed in a hammer mill for a direct comparison with commercial knives made of casehardened low carbon steel.

4.3.1 Laboratory wear test

High-strength ausferritic steels have a number of desirable mechanical properties and have thus been viewed as a candidate material for heavy wear applications. Examinations of the sliding wear resistance of high silicon ausferritic steels have shown positive results [29, 30]. The sliding wear resistance in this research was examined in laboratory under controlled conditions using a Cameron plint reciprocating test rig. The Cameron Plint test rig is very versatile and can be used in both lubricated and dry conditions, with various test geometries, and the test can easily be repeated for comparisons with other materials.

The test was performed in dry condition with a point contact. With dry conditions a large amount of wear can be expected, which makes measurements that are necessary to determine the wear resistance and comparison between different samples easier. With a point contact it is possible to obtain high contact pressure, which is of particular interest for the bainitic steel to see if martensitic transformation of the retained austenite due to mechanical deformation would occur. Martensite formation would increase the hardness where wear occurs, and thus form a hard wear resistant layer. As counterpart, a RB-6/G20W SKF ball bearing ball with a hardness of 860 HV0.3 was used. Two different loads, 20 and 300N, were used to examine wear behaviour differed under different load conditions. The pressure between a cylinder and a plate can be determined with the equation for Hertzian line contacts (equation 12), giving an initial maximum contact pressure of 1.7 GPa for 20 N and 4.3 GPa for 300 N normal load. The pressure then quickly decreases as wear takes place and the contact area increases.

\[
P = 0.388 \left( \frac{F_N E^2}{R^2} \right)^{\frac{1}{3}} \quad \text{eq. 12}
\]

\[\begin{align*}
  f & \quad \text{Maximum normal pressure} \\
  F_N & \quad \text{Normal load} \\
  E & \quad \text{Elastic modulus} \\
  l & \quad \text{Length of cylinder} \\
  R & \quad \text{Radius of cylinder}
\end{align*}\]

A comparison of the wear behaviour was made between silicon alloyed steel with ausferritic microstructure and steels with other microstructures according to Table 4. A comparison with boron steels were of particular interest since they are commonly used in wear applications where a through hardness is desirable. The 0.56C-1.9Si steel was heat treated in different ways to examine how different heat treatments and microstructures influence the wear properties. A case hardened 0.1C-low alloy steel that
is being used in cutting applications where it could possibly be replaced with a through hardened ausferritic steel was wear tested on the casehardened surface as well as in the centre.

Table 4 Specimens used in wear test, ordered by hardness. 1)0.25C-boron steel, 2)0.1C-low alloy steel, 3)0.56C-1.9Si-steel, 4)0.25C-low alloy steel, 5)0.18C-boron steel, 6)0.17C-boron steel, 7)0.26C-boron steel 8)0.27C-boron steel 9)0.3C-boron steel

<table>
<thead>
<tr>
<th>Steel &amp; treatment</th>
<th>Microstructure</th>
<th>Hardness [HV0.3]</th>
</tr>
</thead>
<tbody>
<tr>
<td>a 1)Boron steel, spherodized</td>
<td>spherodite</td>
<td>180</td>
</tr>
<tr>
<td>b 2)C-steel, quenched</td>
<td>martensite</td>
<td>260</td>
</tr>
<tr>
<td>c 3)Si-steel, spherodized</td>
<td>spherodite</td>
<td>297</td>
</tr>
<tr>
<td>d 4)C-steel, tough hardened</td>
<td>martensite</td>
<td>344</td>
</tr>
<tr>
<td>e 3)Si-steel, hot rolled</td>
<td>pearlite</td>
<td>349</td>
</tr>
<tr>
<td>f 3)Ausferrite, austemp. 300°C</td>
<td>ausferrite</td>
<td>461</td>
</tr>
<tr>
<td>g 5)Boron steel quenched,</td>
<td>martensite</td>
<td>479</td>
</tr>
<tr>
<td>h 3)Ausferrite, austemp. 340°C</td>
<td>ausferrite</td>
<td>486</td>
</tr>
<tr>
<td>i 3)Ausferrite, austemp. 320°C</td>
<td>ausferrite</td>
<td>498</td>
</tr>
<tr>
<td>j 6)Boron steel, quenched</td>
<td>martensite</td>
<td>505</td>
</tr>
<tr>
<td>k 7)Boron steel, quenched</td>
<td>martensite</td>
<td>510</td>
</tr>
<tr>
<td>l 8)Boron steel, quenched</td>
<td>martensite</td>
<td>530</td>
</tr>
<tr>
<td>m 1)Boron steel, quenched</td>
<td>martensite</td>
<td>560</td>
</tr>
<tr>
<td>n 9)Boron steel, quenched</td>
<td>martensite</td>
<td>590</td>
</tr>
<tr>
<td>o 3)Si-steel, tempered 190°C</td>
<td>martensite</td>
<td>840</td>
</tr>
<tr>
<td>p 2)C-steel, case hardened</td>
<td>martensite</td>
<td>912</td>
</tr>
</tbody>
</table>

The wear behaviour of the ausferritic steel proved to be consistently better compared to the boron steel of similar hardness for both 20 N and 300 N load (Figure 14, 15). One interesting phenomenon is that there is no unambiguous relationship between hardness and wear. The best wear results were obtained from the specimens with a hardness similar to the hardness of the bearing ball, but for specimens with lower hardness a large scatter in wear was found.
It seems that good wear properties are obtained when a combination of sufficient hardness and ductility is achieved and that the influence of hardness becomes more important as the load increases. This is well illustrated by the test specimen with the lowest hardness, a spherodized boron steel, that exhibits a wear comparable to the much harder hot rolled, austempered and quenched Si-steel specimens for the lower load. For the higher load, however, the spherodized boron steel wore off at a rate more than ten times higher compared to the silicon steels. The width of the wear track, caused by the reciprocating ball on the flat specimens, at the end of the wear test, is approximately 1 mm wide, giving a contact area of approximately 0.75 mm$^2$. For a 20 N load, this means...
the contact pressure is reduced from the initial 1.7 GPa down to roughly 25. Similarly, for the 300 N load the pressure is reduced from 4.3 GPa down to ~ 400 MPa. This also means that for the 300 N load, the pressure exceeded the ultimate tensile strength of the spherodized boron steel and could explain the dramatic change in wear. The wear of the hot rolled silicon steel is slightly higher compared to the ausferritic steel. Both steels have the ability of plastic deformation before fracture and have a very fine microstructure and difference in wear can therefore be derived to the difference in hardness.

The ausferritic steels consistently presented better wear results in comparison to the quenched boron steels of similar hardness. Microscope investigation of the wear tracks show that large shear deformation has occurred before formation of wear debris in all samples but the degree of deformation differs between the samples. The martensitic structures consistently show less deformation and the wear track is rougher compared to the ausferritic samples where the dimples from wear debris are smeared over. A more ductile behaviour of the ausferritic steels thus seems to have a prevented wear debris to break lose and thereby lowered the amount of wear.

The wear of the ausferritic specimens showed indications of contact fatigue (Figure 16). Beneath the worn surface, horizontal cracks have formed. It seems the wear debris has formed from flat particles that break lose as the cracks reach the surface. The uppermost layer under the wear scar is heavily deformed. The visibly deformed layer is about 20 μm thick, whereas the uppermost 5 μm is deformed to such extent that the ausferritic structure no longer can be identified. This layer is usually assumed to be composed of metal that has been transferred back and forth between the ball and the plate along with various oxides which have been intimately incorporated [29]. When the oxide layer reaches a critical thickness, the stress in the interface with the underlying metal reach a critical value resulting in flaking [31]. It is not fully understood why this phenomenon is only present in the ausferritic specimens and further investigation is needed to clarify its origin and influence on the wear progress.
A significant increase in hardness was found in the wear track and the deformed layer beneath the wear scar. The hardness of sample h, austempered at 340°C, had increased from 486 HV$_{0.3}$ to 690 HV$_{0.3}$ after wear test at 300 N. The hardness increase corresponds to the degree of deformation, and at 40 μm into the material, no significant increase in hardness is detected (Figure 17). After the wear test at 20 N load, the hardness increase was found to be 85 Vickers with a similar depth of deformation.
Almost no wear of the ball bearing balls could be measured when worn against surfaces of the ausferritic samples at 300 N load whereas the martensitic boron steels caused a much higher wear on the bearing balls, as can be seen in Figure 18. A possible explanation for this could be found at microstructural level, where the silicon prevents hard abrasive carbides to form in the ausferrite.

![Figure 18](image)

**Figure 18** Wear of the ball bearing ball after wear testing under 300 N load. The wear of the balls was found to be negligible after wear test of the ausferritic specimens. This could be explained by the fact that no hard, abrasive carbides are present after austempering.

The results from wear testing show that the ausferritic steel possesses large potential in finding applications where high strength and toughness is needed. Further testing of ausferritic steels with higher carbon content and higher hardness would be of interest, as well as evaluation of the surface fatigue resistance.

### 4.3.2 Hammer mill test

The wear resistance test in laboratory environment was followed up by a comparative test in production equipment at Luleå Energi. Four knives of high silicon steel were manufactured and heat treated by austenitization followed by austempering at 340°C for two hours at Componenta in Karkkila, Finland. The austempered knives were mounted in the hammer mill together with commercial knives for a comparative test. The commercial knives are made of case hardened low carbon steel with a surface hardness of 900 Vickers. The hardness gradually decreased down to 260 Vickers in the centre of the specimen. The austempered high silicon steel has a through hardness of 485 Vickers, but due to decarburization the surface hardness was lowered down to 265 Vickers, as illustrated in Figure 19.
The hammer mill was regularly stopped for visual inspection of the wear progress. It was expected that the hard surface layer of the case hardened steel would prevent wear, resulting in an initially higher wear on the austempered knives. However, the difference in wear was found to be small and differ more between the individual knives than between the different steels. It should be noted that visual inspections are not very accurate in estimating the actual wear. Regular weight measurements would thus have been needed in order to determine the influence of a hard surface layer on the wear progress. A quantitative determination of the wear was made by weight loss measurements when the knives were demounted to be turned and replaced. The results of the weight measurements can be seen in Figure 20. It was found that the wear rate of the ausferritic knives is equal or less compared to the casehardened knives. Even though the hardness difference is large, the ausferritic steel performs as well as the traditional knives. Unfortunately, the decarburized layer of the austempered knives may have influenced the wear.

**Figure 20** Measured weight loss of knives in hammer mill. First measurement was taken after seven weeks when first two corners were worn out and second measurement after 13 weeks when all four corners were worn out and knives replaced.
The knives were found to have been exposed to impacts from larger debris during operation, resulting in indents on the surfaces. Fortunately, all of the knives had withstood the impacts and none of the knives had ruptured. The impact toughness of the austempered is thus sufficient for the investigated application.
5 Conclusions

Austempering of a steel with the chemical composition; Fe-0.56C-1.9Si-0.84Mn-0.19Cr-0.11Ni (wt%), for two hours at temperatures between 295°C and 340°C results in an ausferritic microstructure consisting of fine laths of ferrite separated by thin films of carbon enriched austenite. No carbides were detected. The austenite was found to be very stable and presented high resistance to martensite transformation even when exposed to cryogenic heat treatment and mechanical deformation.

The mechanical properties were characterized and compared to the properties of the steel in its as-rolled condition. The toughness, hardness, yield- and ultimate strength were significantly improved and the ductility was preserved after austempering at 340°C. Austempering resulted in a homogeneous microstructure with a constant through-section hardness. A high impact toughness at low temperatures in combination with high strength vouch for numerous possibilities to utilize the austempered high silicon steel, in both indoor and outdoor environment, when high performance is desired.

The wear resistance of austempered high silicon steel is better compared to martensitic boron steel of similar hardness under dry sliding conditions. This can be attributed to the carbide free microstructure, which not only presented high wear resistance, but also proved to cause less wear on the counterpart compared to other microstructures during wear testing. This phenomenon, with low wear on the counterpart, should be of interest in applications that rely on fine tolerances but suffer from an insufficient lubrication regime.

In a field test for the steel's potential application as knives in a hammer mill used for the fractioning of sawmill dust for wood pellet production, it was found that austempered high silicon steel can be used successfully. Hence, it can be concluded that the austempered high silicon steel also is suitable to be used in other applications, e.g. earth moving equipment or conveyor structures, where high wear resistance and impact toughness is required.
6 Future work

This work showed that the ausferritic steel has good potential for numerous industrial applications, however there are several areas that require further development to allow this potential to be fully utilized. Only one type of steel has been investigated in this research, therefore it would be of interest to investigate steels with higher carbon content and other alloying additions in order to tailor the steel properties to meet different application requirements.

Surface hardening could be examined as a method to further improve the wear resistance of the ausferritic steel.

The wear testing in this research has been performed under sliding conditions. In order to better determine the wear resistance, other wear phenomenons, such as erosive wear and surface fatigue, should also be studied. Especially the surface fatigue under non-lubricated sliding wear conditions would be of interest as a complement the results of this research.

The heat treatment used in this research was carried out by austenitization followed by isothermal transformation in a salt bath. The possibility of continuous cooling should be looked into, as well as other possible methods of isothermal transformation.

Lastly, the properties of high silicon steel can find use in other applications. In this work the wear resistance has been emphasized, but the ultra high strength properties that can be achieved by austempering should be applicable to numerous other applications and should be looked at.
References


[22] ASM Handbooks Online, volume 10, X-ray powder diffraction


