Control of weldability
Research leading to the development of two new quenched and tempered tool steels.

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Dissertation

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Preface

After obtaining the MSc Met engr the present author worked with projects on weldability research at the Swedish Institute of Metals Research (SIMR) in Stockholm. At the same time Dr’s courses and classes were taken at KTH on welding technology. After five years of basic welding technological research at the SIMR the field of product development in the steel industry tempted and the author came to work with product development of quenched and tempered steels with rather high hardnesses/strength levels such as abrasion resistant steels, armour plate and the like. The similarity between the work regarding HAZ hardenability and the hardenability of quenched and tempered steels may at a glance not be obvious, but the HAZ in a weldment is indeed a quencher in miniature and the principles of materials behaviour in the HAZ in a weldment and the behaviour of a plate being quenched in a quencher are basically the same. What was gained in knowledge during the early years was to be the leading star for the product development of the quenched and tempered steels.

Finally, two new tool steels have been developed with low carbon contents for being tool steels although high in carbon for weldable steels. Here the principles for hardenability have been used both in the sense of producing a quench and tempered steel according to rules formalised by Grossman and his colleagues and with use of the hardenability equations set up first by Dearden and O’Neill. These latter were solely developed to forecast the behaviour of the steel when electric arc welding with the coated electrode. It appears that the two different approaches were developed in the years around 1940 quite independently. In the present thesis it is shown how the two different paradigms come to meet in steel development of tool steels intended for moulds for plastics and rubber processing.

A welding wire, of the same chemical composition as the newly developed tool steels, was also developed for autogeneous welding of the mould steels, as no commercial wire fit for repair welding of these materials could be found on the market. Using TIG-welding this filler wire also turned out to produce a weld metal which can match the mechanical properties of super high strength steels having yield strengths of not less than 900 MPa.

This thesis consists of an introductory part and the following papers:


Paper K  Larsson C., Hansson P., Kihlmark P. 'Toughness of HRC 33/44 Tool Steels Weldment with Yield Strength 1100 MPa’. To be published.


Oxelösund, October 2004

Per Hansson
Abstract

The understanding of the hardenability is important in steel development with respect to weldability of steels as well as to the design of quenched and tempered steels. The common way to judge if steel is suitability to welding is the use of a carbon equivalent, which reflects the alloy content to the hardenability of the heat affected zone (HAZ). Most common of these equivalents is the IIW carbon equivalent which has been in use for decades. However, this is an empirical equivalent, developed for CMn steels, and it doesn’t say anything about the mechanical properties obtained in the HAZ. As a guideline a maximum HAZ hardness of 350 HV is normally put to avoid cold cracking in the HAZ. Numerous carbon equivalents have been published during the years, but few have been accepted and used. A drawback of all these equivalents is that they have been evaluated for a set of chemical compositions which may be quite narrow, or that they have been evaluated on steels produced having other cleanliness, residual element levels etc. as is common in modern steelmaking.

The hardenability of steel designated to quenching and tempering is normally evaluated using either Jominy testing or using a calculation using the Grossmann formalism.

As the hardenability described in the carbon equivalents used in welding and the hardenability of quenched and tempered steels using the Grossmann formalism are two different descriptions of the same phenomenon the aim of this thesis is to link these two formalisms together and use the knowledge from the HAZ hardenability to design quenched and tempered tool steels. The goal is to produce such steel grades having a much better weldability as compared with standard grades available at the market.

In this work the hardenability in the heat affected zone (HAZ) in weldments in high strength low alloyed steels (HSLA steels) having a yield strength of 350 MPa. This part of the study includes microstructural studies of the HAZ and the influence of microalloying elements on the phase transformations which take place during cooling from the peak temperature of the welding cycle. These results have been linked to the mechanical properties of the HAZs.

In the second part of this thesis the development of tool steel grades with respect to optimize both hardenability with respect to:

- matrix hardenability (i.e the Grossmann approach)

- an enhanced weldability of such new steels due to a lower alloying content than is usual in such grades.

These new steels also possess a superior Machinability as compared ton the common standard grades W.nr 1.2311 and W.nr 1.2312.

Key words: Constructional steels, microalloying, tool steels, heat affected zone, hardenability, microstructure, precipitation, grain size, welding, carbon equivalent, toughness
List of abbreviations and symbols

<table>
<thead>
<tr>
<th>Symbol</th>
<th>Description</th>
</tr>
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<tbody>
<tr>
<td>α</td>
<td>Ferrite</td>
</tr>
<tr>
<td>AC</td>
<td>Bainite, Widmanstätten ferrite</td>
</tr>
<tr>
<td>AF</td>
<td>Acicular ferrite</td>
</tr>
<tr>
<td>CCT</td>
<td>Continuous Cooling Transformation</td>
</tr>
<tr>
<td>CE_{IIW}</td>
<td>Carbon equivalent according to IIW</td>
</tr>
<tr>
<td>CEN</td>
<td>Carbon equivalent according to Yurioka</td>
</tr>
<tr>
<td>CET</td>
<td>Carbon equivalent according to Thyssen Stahl</td>
</tr>
<tr>
<td>CTOD</td>
<td>Crack Tip Opening Displacement</td>
</tr>
<tr>
<td>Di</td>
<td>Grossmann hardenability</td>
</tr>
<tr>
<td>Dic</td>
<td>Grossmann hardenability regarding austenite grain size and carbon content</td>
</tr>
<tr>
<td>dγ</td>
<td>Austenite grain size</td>
</tr>
<tr>
<td>Δt_{8/5}</td>
<td>Cooling time from 800 to 500°C</td>
</tr>
<tr>
<td>Eus</td>
<td>Upper shelf energy at impact testing</td>
</tr>
<tr>
<td>f</td>
<td>Fraction</td>
</tr>
<tr>
<td>F_{B}</td>
<td>Hardenability factor of boron</td>
</tr>
<tr>
<td>γ</td>
<td>Austenite</td>
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<td>HAZ</td>
<td>Heat-affected zone</td>
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<tr>
<td>HBW</td>
<td>Brinell hardness tungsten ball</td>
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<td>HV</td>
<td>Vickers hardness</td>
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<tr>
<td>HRC</td>
<td>Rockwell hardness</td>
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<tr>
<td>HSLA</td>
<td>High Strength Low Alloy</td>
</tr>
<tr>
<td>IIW</td>
<td>International Institute of Welding</td>
</tr>
<tr>
<td>ITT_{27J}</td>
<td>Impact Transition Temperature at 27 J</td>
</tr>
<tr>
<td>K</td>
<td>Solubility product</td>
</tr>
<tr>
<td>MAG</td>
<td>Metal Arc Active Gas</td>
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<tr>
<td>MIG</td>
<td>Metal Arc Inert Gas</td>
</tr>
<tr>
<td>M(σMe)</td>
<td>Molar weight of microalloy</td>
</tr>
<tr>
<td>M(N)</td>
<td>Molar weight of nitrogen</td>
</tr>
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<td>P_{cm}</td>
<td>Carbon equivalent according to Ito Bessyo</td>
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<td>Polygonal ferrite</td>
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<tr>
<td>r_{c}</td>
<td>Precipitate size</td>
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<tr>
<td>R_{0}</td>
<td>Average austenite grain size</td>
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<tr>
<td>TIG</td>
<td>Tungsten Inert Gas</td>
</tr>
<tr>
<td>TMCP</td>
<td>Thermo Mechanical Controlled Processed</td>
</tr>
<tr>
<td>Q</td>
<td>Heat input</td>
</tr>
<tr>
<td>Q&amp;T</td>
<td>Quenched and Tempered</td>
</tr>
<tr>
<td>SAW</td>
<td>Sub-merged Arc Welding</td>
</tr>
<tr>
<td>SEGA</td>
<td>Stora ESAB Gränges AGA research co-operation</td>
</tr>
<tr>
<td>T</td>
<td>Temperature</td>
</tr>
<tr>
<td>wMe^{d}</td>
<td>Weight-% micro-alloying element dissolved in austenite</td>
</tr>
<tr>
<td>wMe^{st}</td>
<td>Weight-% micro-alloying element in steel</td>
</tr>
<tr>
<td>wN^{d}</td>
<td>Weight-% nitrogen or carbon dissolved in austenite</td>
</tr>
<tr>
<td>wN^{st}</td>
<td>Weight-% nitrogen or carbon in steel</td>
</tr>
<tr>
<td>W.nr</td>
<td>Werkstoffnummer, according to German standard</td>
</tr>
<tr>
<td>Y.P</td>
<td>Yield point</td>
</tr>
<tr>
<td>Z</td>
<td>Radii ratio between growing grains to the average grains</td>
</tr>
</tbody>
</table>
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1. Hardenability.

1.1 General

The title of the present thesis is Control of hardenability…, and thus hardenability as a phenomenon requires a short discussion and definition. What is hardenability and how is it measured and defined in the present thesis cases?

Hardenability of steel is the material’s inherent ability to respond to cooling by forming martensite. Since far back in the middle of the 1930s a rather good appreciation of how different elements, like carbon in the first hand, influence the ability to form martensite also in the interior of thick material pieces. Jominy presented his test procedure in 1938, and Grossmann made his first approaches, already in 1942, to what later appeared as the Grossmann formulae for calculating the hardenability (1), a method that although modified is still in use.

In the 1920s the first altogether welded ships were made and also the first welded pressure vessels. The first welded ocean going ships were produced in the early thirties. It was conceived that arc welding with a coated electrode required a low carbon content in the base material in order that there should be no risk for spontaneous cracking in the HAZ of the steel. It was also conceived that other alloy elements like manganese and chromium increased the risk for this type of cracking i.e. they had an influence on cracking similar to that of carbon. Weldability came under discussion and in 1940 a carbon equivalent weldability formula was introduced by Dearden and O’Neill (2), where the relative influence on microstructure/hardness of the different important elements in steel were weighted against the influence of carbon.

It is not altogether clear whether the metallurgists working with mild steels were influenced by those working on the high strength steels of the time. But it is very likely that physical metallurgists stemmed from the same universities, and that ideas floated freely between the structural steel metallurgists and the metallurgists working with high strength engineering steels and with the high strength tubing as used for structural purposes for aircraft fuselages before aluminium took over. The low alloy tubing and plate also appeared necessary for the ammonium synthesis production later standardised in terms of the Nelson diagrams, showing the influence of hydrogen partial pressure and temperature, and requirements of CrMo-alloying. Also for steam boilers the low alloy CrMo steels were used before welding became applied, i.e the use of steel physical metallurgy for general purposes matured during the late thirties and forties.

The Grossman formulae can easily be transformed to a carbon equivalent similar to that of Dearden and O’Neill. However, their formula appeared before the Grossman factors were finally published.

Today Dearden and O’Neill’s carbon equivalent is only used when calculating the risk for hydrogen cracking on welding, and this is now an EN standard (EN 10 113). Three more carbon equivalents for welding have since become standardized methods, but ISO has given up hope on an ISO standard by now.
The scope of weldability is not only a scope of resistance to hydrogen induced spontaneous cracking. It also is a question of toughness and to some extent also of static strength of the weld metal and the heat affected zone. Furthermore, the toughness can be seen as a critical measured CTOD value, a Charpy toughness at a certain temperature, a Charpy-V transition temperature, a Kc value or the like. Finally it can be related to a critical crack size and it appears that almost always the weld metal or the heat affected zone is the critical region in a welded joint, even if the weld is thoroughly stress relief heat treated. In the present thesis an example of such fitness for purpose calculations based on welding with a matching wire is given. The matching wire has also been tested with high strength Q&T steel with good results as regards toughness, both with MIG/MAG- and TIG-welding.

It is self-evident that neither the Dearden&O’Neill approach nor the Grossman number says anything about toughness which has to be studied separately.

In the present thesis the research on toughness and strength of weldments has provided knowledge, later used in formulating two new low alloyed tool steels. The hardenability in terms of the Grossman factor must be sufficient for the manufacture of the steel at the same time that the hardenability in terms of the CEIIW-carbon equivalent should be as low as possible to make possible repair welding and also structural welding in some cases.

1.2 The transformation from Grossman hardenability to CEIIW

In this thesis the hardenability of HAZ in welds and the hardenability of quenched and tempered steels have been studied. An approach has been made where it is shown that the Grossman hardenability can be transformed to something similar to that of Dearden and O’Neill (today the CEIIW formula). This demonstrates that the two approaches to hardenability reflect very much the same influence on the CCT diagram:

Consider a grain size of ASTM 1, then, according to Moser and Legat (3), the function describing the influence of carbon content and austenite grain size on the hardenability is 32.5·C₁/₂. Assume a linear function from zero to 0.2% C where the inclination is determined by a straight line from the 0.1% C-value to the corresponding 0.2% C as calculated according to above. The hardenability depth with respect to carbon content and austenite grain size, i.e the Dic value then can be written:

\[ \text{Dic} = D_{io} + k \cdot C \]
\[ D_{io} = (6.01 \cdot C + 42.6 \cdot C) \times (1 + 0.40 \cdot \text{Si}) \times (1 + 1.21 \cdot \text{Mn}) \times (1 + 1.13 \cdot \text{Cr}) \times (1 + 0.47 \cdot \text{Ni}) \times (1 + 2.275 \cdot \text{Mo}) \]

Dividing both sides by 6 and then taking the natural logarithm of both sides gives,

\[ \ln(D_{ic}/6) = \ln(1 + 7.1 \cdot C) + \ln(1 + 0.40 \cdot \text{Si}) + \ln(1 + 1.21 \cdot \text{Mn}) + \ln(1 + 1.13 \cdot \text{Cr}) + \ln(1 + 0.47 \cdot \text{Ni}) + \ln(1 + 2.275 \cdot \text{Mo}) \]
Using a series expansion of $\ln(1+x) = x$ for small $x$-values and neglecting higher terms gives:

$$\ln(\frac{D_i}{6}) = 7.1 \cdot C + 0.40 \cdot Si + 1.21 \cdot Mn + 1.13 \cdot Cr + 0.47 \cdot Ni + 2.27 \cdot Mo$$

Division by 7.1 gives

$$\frac{\ln(\frac{D_i}{6})}{7.1} = C + \frac{Si}{18} + \frac{Mn}{5.87} + \frac{Cr}{6.28} + \frac{Ni}{15} + \frac{Mo}{3.13}$$

which extremely close to the Dearden and O’Neill formula, elements in weight-%;

$$CE = C + \frac{Mn}{6} + \frac{(Cr + Mo + V)}{5} + \frac{(Cu + Ni)}{15}.$$  

2. The carbon equivalent with respect to weldability

Numerous carbon equivalents exist to assess the weldability of a steel in terms of hardenability and risk for cracking. Since decades it is well known that the cold cracking susceptibility is not possible to link directly to the carbon equivalent since an influence exists from the

- microstructure,
- hydrogen content, and
- restraint condition.

A possible method to evaluate the cold cracking susceptibility is to correlate the hardness in the heat-affected-zone (HAZ) and the actual carbon equivalent. The easiest way to describe the hardness in the HAZ is to adopt a suitable carbon equivalent such as the IIW-equivalent, $CE_{IIW}$, which only corresponds to the influence of the major alloying elements in the steel, see Equation 1.

$$CE_{IIW} = C + \frac{Mn}{6} + \frac{(Cr + Mo + V)}{5} + \frac{(Cu + Ni)}{15} \quad (Eq 1)$$

Alloying levels in Eq 1 are given in weight-%. This equivalent was developed to describe the hardenability in C- and CMn-steels. Commonly, a hardness limit of maximum 350 HV in the HAZ is required. This can be traced back to the observations reported by Dearden and O’Neill (2), who observed that no cracks occurred in HAZs having hardnesses lower than 350 HV. In other words, a minimum cooling time, $\Delta t_{8/5}$, has to be established with respect to the actual plate chemical composition and weld method used, to minimize the amount of martensite in the HAZ.

Another well-known carbon equivalent, $Pcm$, has been proposed by Ito et al (4) to describe the cold-cracking susceptibility in high strength steels, produced as pearlite reduced thermo-mechanically controlled rolled steels. It is given in Equation 2, where all elements are given in weight-%.

$$Pcm = C + \frac{Si}{30} + \frac{(Mn + Cu + Cr)}{20} + \frac{Ni}{60} + \frac{Mo}{15} + \frac{V}{10} + 5B \quad (Eq 2)$$
The equivalent has been evaluated from high strength steels having chemical compositions as given below.

Carbon 0.07 – 0.22
Silicon max. 0.6
Manganese 0.4 – 1.4
Chromium max. 1.2
Copper max. 0.5
Molybdenum max. 0.7
Nickel max. 1.2
Vanadium max. 0.12
Boron max. 0.005

The authors (4) correlated the cold-cracking susceptibility, evaluated as the portion of cracks in a section in a Y-groove restraint cracking test, with the plate thickness and amount of diffusible hydrogen in the weld metal. The Pcm value was correlated to this test as, see Equation 3,

\[ P_c = P_{cm} + \frac{t}{600} + \frac{H}{60} \]  (Eq 3)

where \( t \) = plate thickness, in mm, and \( H \) = diffusible weld metal hydrogen content, in ml/100 g weld metal. Equation 3 is valid for plate thicknesses 19 – 50 mm and diffusible hydrogen contents between 1.0 and 5.0 ml/100g weld metal.

Yurioka et al (5) have grouped carbon equivalents into two groups. In the first group carbon has a larger influence than the other elements on the carbon equivalent. The Pcm-equivalent belongs to this group. In the second group other alloying elements than carbon have a larger influence with respect to the carbon content, and an example of these equivalents is the CEIIW. Furthermore, they have shown that the cold-cracking susceptibility can best be described by the Pcm when the carbon content is below 0.16%, and by the Ew with carbon contents from 0.16%. This has been summarized in the CEN-equation proposed by the authors, see Equation 4.

\[ CEN = C + A(C) \times \left[ \frac{S}{24} + \frac{Mn}{6} + \frac{Cu}{15} + \frac{Ni}{20} + \frac{(Cr+Mo+V+Nb)}{5} + 5B \right] \]  (Eq 4)

Where \( A(C) = 0.75 + 0.25 \tanh[20(C-0.12)] \)  (Eq 5)

It is evident that these proposed carbon equivalents, have not been intended to use when judging the weldability in QT-steels. Thyssen Stahl has published a carbon equivalent also suitable to QT-steels, see Equation 6.

\[ CET = C + \frac{(Mn+Mo)}{10} + \frac{(Cr+Cu)}{20} + \frac{Ni}{40} \]  (Eq 6)

All elements in weight-%. This equivalent is reported to be valid for steels having chemical compositions as given below.
Carbon 0.05 – 0.32
Silicon max. 0.8
Manganese 0.5 – 1.9
Chromium max. 1.5
Copper max. 0.7
Molybdenum max. 0.75
Nickel max. 2.5
Niobium max. 0.06
Titanium max. 0.12
Vanadium max. 0.18
Boron max. 0.005

When welding quenched and tempered steels there is a demand that the HAZ shall not be too wide and too soft, i.e. a maximum cooling time $\Delta t_{8/5}$ has to be limited.

ISO has compared the required preheat temperature as calculated with the four methods, and in some cases one method is more conservative than the others. This differs with heat input, hydrogen content, plate thickness and carbon equivalent.

A drawback of all carbon equivalents proposed is that they have been established at different times for steels having different residual element levels.

3. Hardenability calculation of steels

When designing a steel dedicated to heat treatment it is necessary to describe its hardenability, i.e. the ability of a steel to transform into martensite and/or bainite when cooling from the austenite phase. Numerous hardenability descriptions have been suggested, but it is the Grossmann (1) formalism which is normally used when describing a steel's hardenability. These descriptions are based on slow heating rates to the austenitizing temperature, long soaking times when this temperature has been reached thereby giving a fully transformation into austenite. In a HAZ the heating rate is much higher and the soaking time at the maximum temperature is also very short thereby affecting the transformation into austenite. Also the quenching rates are normally lower when quenching a steel as compared to the cooling rate of a HAZ.

Grossmann, however, based his original study (1) on a quite limited range of chemical compositions and he suggested that the hardenability factors could be linearly extrapolated towards higher alloying contents. A benefit when using the Grossmann formalism is that it can be linked to the Jominy test. Moser and Legat (3) have shown that when the alloying contents increase, the multiplication factors show a logarithmic increase. Just (6) has presented a model where he predicts the hardness in a steel as calculated hardness at a given Jominy distance. A major drawback of the methods given above is that they do not take into account the influence of austenitizing temperature and soaking time at this temperature. These methods may not give a proper description of steel hardenability since all precipitates may not be in solution in
austenite when quenching the steel, i.e. the kinetics of precipitate dissolution are neglected. Thereby, a lower hardenability than predicted according to the nominal chemical composition would be expected.

During the last decades boron has been more attractive due to its powerful influence on hardenability when alloyed in small additions. A boron addition which gives about 20 ppm free boron during austenitizing doubles or triples the hardenability, i.e. a value of the boron factor, $F_B$, about 2 – 3. This is utilized in the modern microalloyed steels giving the possibility to add such a small amount of boron and reduce the additions of more expensive alloying elements such as molybdenum and nickel. However, a complication is that hardenability in modern boron containing steels is difficult to describe because of numerous reasons of which the three most important are:

- determination of free boron in austenite
- uncertainty of the effective hardenability factor from boron
- uncertainty of what is the optimum free boron in austenite with respect to hardenability.

Figure 1 (7) shows the large variation in boron factor depending on boron content in a steel containing 0.2% C, 0.8% Mn and 0.6% Mo.

![Figure 1. Influence of boron content on hardenability. (7)](image-url)
Rahrer and Armstrong (8) have given the following relationship to describe the hardenability factor of boron in unalloyed and low-alloyed steels:

\[ F_B = 1 + 1.5 \cdot (0.9 \text{-weight}\% \text{C}) \]  
(Eq 7)

The reduction of the boron factor with increasing carbon content is dependent on a precipitation of boron carbides. However, when regarding the influence of boron on hardenability in steels notice must also be taken to the protection of boron against nitrogen due to the strong affinity between boron and nitrogen. This protection is commonly made utilizing titanium and/or aluminium as these are stronger nitride formers than boron.

4. Steel development

In modern steel development, weldability is of major interest since welding is the most common method to join different components made of steel. When judging whether a steel is suitable for welding or not one has to consider a lot of different aspects regarding the steel itself, the filler material, the welding method used and the properties of the weldment with regard to its intended functionality. This paper will discuss weldability aspects of CMn-steels, HSLA-steels, Thermo Mechanical Controlled Processed (TMCP) steels and also low alloyed Quenched and Tempered steels (Q&T-steels).

The trend towards higher and higher yield strengths in structural steels, higher hardnesses in wear resistant steels, better repair weldability in tool steels have resulted in the usage of different steelmaking and hot working routes when fabricating a steel product (plate, beam, forging etc.). In the following, is a discussion of the development and production of heavy plates.

As is well-known, when designing a steel a metallurgist has five strengthening mechanisms in a steel to bother about;

- Solid solution strengthening
- Precipitate strengthening
- Grain size strengthening
- Deformation strengthening
- Pierls-Nabarro strengthening

Of these mechanisms only the grain size strengthening gives both an increase in strength and in toughness. The others increases strength and decreases toughness. Obviously the development is focussed on producing as fine-grained steels as possible using as low other alloying elements as possible for the strength level desired.

Table 1 gives different steel grades, having alloying contents resulting in more or less equal carbon equivalents in plate thicknesses 10-20 mm, grouped with respect to the production processing routes used.
TMCP-steels normally have yield points from 420 to 500 MPa while Q&T-steels have yield points from 700 to 1100 MPa. Today, wear resistant steels are produced having nominal hardnesses from 400 up to 600 HBW.

<table>
<thead>
<tr>
<th>Element</th>
<th>S235</th>
<th>S355</th>
<th>S460M</th>
<th>S690Q</th>
<th>S890Q</th>
<th>WX1100</th>
<th>AR400</th>
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<tr>
<td>C</td>
<td>0.11</td>
<td>0.13</td>
<td>0.085</td>
<td>0.13</td>
<td>0.17</td>
<td>0.19</td>
<td>0.12</td>
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<tr>
<td>Mn</td>
<td>1.13</td>
<td>1.5</td>
<td>1.55</td>
<td>1.0</td>
<td>1.4</td>
<td>0.85</td>
<td>1.4</td>
</tr>
<tr>
<td>Cr</td>
<td>-</td>
<td>-</td>
<td>0.25</td>
<td>0.25</td>
<td>0.6</td>
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<tr>
<td>Mo</td>
<td>-</td>
<td>-</td>
<td>0.15</td>
<td>0.5</td>
<td>0.6</td>
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<td>-</td>
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<tr>
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<td>-</td>
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<td>-</td>
<td>-</td>
<td>-</td>
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<tr>
<td>Ni</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>1.9</td>
<td>-</td>
</tr>
<tr>
<td>CEIIW</td>
<td>0.30</td>
<td>0.39</td>
<td>0.37</td>
<td>0.40</td>
<td>0.56</td>
<td>0.71</td>
<td>0.37</td>
</tr>
<tr>
<td>Prod. route</td>
<td>As</td>
<td>As</td>
<td>TMCP</td>
<td>Q&amp;T</td>
<td>Q&amp;T</td>
<td>Q&amp;T</td>
<td>Q</td>
</tr>
<tr>
<td>$R_{p0.2}$ (MPa)</td>
<td>≥ 235</td>
<td>≥ 355</td>
<td>≥ 460</td>
<td>≥ 700</td>
<td>≥ 900</td>
<td>≥ 1100</td>
<td></td>
</tr>
<tr>
<td>HBW Aim</td>
<td>400</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

Table 1. Typical chemical compositions for plate thicknesses 10-20 mm. Alloying contents in weight-%. Carbon equivalent according to CEIIW.

As is demonstrated above, structural steels with guaranteed yield strengths up to 700 MPa (S690Q) and 400 HBW abrasive resistant steels (AR400), in plate thicknesses up to 20 mm, can be produced having a carbon equivalent less than or equal to around 0.40, i.e they need no preheat when welded in combined thicknesses up to 40 mm if low-hydrogen, < 5 ml/100g weld metal, welding processes are used.

Today, modern steelmaking practice produces steels which have far lower impurity levels, lower hydrogen and nitrogen levels as compared to the steels produced at the time when the CEIIW-equivalent was developed. Table 2 gives an example of the differences in steels produced today and at the time the CEIIW was established.

<table>
<thead>
<tr>
<th>Element</th>
<th>‘Old process route’ (1950/1960’s)</th>
<th>‘Modern steels’ (1990’s)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Sulphur</td>
<td>100 – 300</td>
<td>50 - 80</td>
</tr>
<tr>
<td>Phosphorus</td>
<td>150 - 300</td>
<td>80 - 140</td>
</tr>
<tr>
<td>Hydrogen</td>
<td>4 - 6</td>
<td>3 - 5</td>
</tr>
<tr>
<td>Nitrogen</td>
<td>80 – 150</td>
<td>&lt; 60</td>
</tr>
<tr>
<td>Oxygen</td>
<td>60 - 80</td>
<td>&lt; 10 to 12</td>
</tr>
</tbody>
</table>

Table 2. Typical impurity levels. All elements in ppm.
In the 1950’s the development of microalloyed steels began, and these grades are nowadays well-known as the HSLA-steels, showing typical yield strengths of 300-350 MPa in combination with a good toughness and an excellent weldability. HSLA steels contain small additions of titanium, vanadium and/or niobium. The increase in strength, as well as in toughness, relates to the fine-grain effect that is possible to achieve via small titanium nitrides hindering austenite grain growth in combination with precipitation of niobium carbonitrides as well as vanadium carbonitrides.

In Sweden, Oxelösunds Jernverk, nowadays SSAB Oxelösund AB, took a major part of the development of niobium-microalloyed HSLA-steels via the work of De Kazcinzy (see Ref. 9) in the early 1960’s. De Kazcinzy studied the possibilities of utilizing niobium as a microalloying element and developed 350 MPa line-pipe-steel grades which were produced at Oxelösunds Jernverk.

To further increase the strength level in a steel the plate can be processed via hot rolling in a temperature interval where no recrystallization takes place between the passes (Thermo Mechanical Controlled Processing, TMCP). These steels have an austenitic microstructure which consists of very elongated and thin grains. The conditioned austenite also consists of deformation bands, and twins boundaries. During cooling after the final rolling pass a very fine grained ferritic microstructure is achieved after the phase transformation $\gamma \rightarrow \alpha$. To control the recrystallization stop when rolling in the austenitic regime, niobium is commonly used. Depending on the carbon, free nitrogen and niobium contents, one aims to achieve a niobium carbide/carbonitride precipitation on sub-grains/grain boundaries/twin boundaries/other defects resulting in a retardation or stop of austenite recrystallization. Normally the recrystallization stop temperatures are around 950-1025°C in steels for heavy plate rolling. The niobium carbide precipitation rate in recrystallized austenite is however sluggish. More niobium carbonitrides can also be precipitated at the austenite-ferrite transformation.

An example of these grades is S460M, given in Table 1. These grades contain very low alloying contents, where the main focus on the alloy concept is the microalloying which results in steels having excellent weldabilities, typical CEIW’s within the range 0.37 – 0.38. When judging the weldability of the TMCP steels it is obvious that the Pcm equivalent, which is developed for low alloy contents, would be the best suited, which also has been pointed out by Yurioka (5) already in the 1980’s.

When even higher strength levels are required (yield strengths from 700 MPa), or when wear resistance is required a further increase in alloying content in combination with TMCP processing is not a suitable production route as it would lead to a too high alloying content from the viewpoint of weldability.

The next step in processing is then to quench and temper the actual plate. These grades (Q&T-steels) have microstructures consisting of tempered lath martensite, and show high strength levels, yield strengths from 700 – 1100 MPa, in combination with good toughness.
In the 1990’s the author of this thesis was involved in the development of the HARDOX abrasive wear resistant steels. A description of the weldability of these grades is given in Paper I. In 2000 it was decided that the possibilities of developing and producing tool steels, dedicated to plastic mould production, should be evaluated at SSAB Oxelösund. The requirements was that:

- the new steel grades shall be produced using the production facilities in Oxelösund
- the steels shall be delivered in quenched and tempered condition
- the steels shall have low alloy contents, enabling an easy repair welding of moulds, as compared to other steel grades on the market.

Looking deeper into the question, and after taking out a property profile, which will be discussed in Chapter 9, on the tool steel it was found that completely new steels had to be formulated, with the author being responsible for the steel development.

Due to the enormous span in chemical composition of tool steels on the market, from water-quenched through oil-quenched to air-hardening ones, it is obvious that these grades have been developed regarding maximum hardenability and no respect has been paid to their weldability. Mainly because when welding the grades of interest, it concerns repair welding of quite small details where proper preheat and slow cooling after welding is very well controlled. Examples of chemical compositions of some frequently used tool steels used are given in Table 3. The W.nr 1.XXXX refers to the frequently used German steel standard.

<table>
<thead>
<tr>
<th>Steel</th>
<th>Application</th>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>Cr</th>
<th>Ni</th>
<th>Mo</th>
<th>V</th>
<th>CE_{IIW}</th>
</tr>
</thead>
<tbody>
<tr>
<td>W.nr 1.2738</td>
<td>Plastic mould steel</td>
<td>0.40</td>
<td>0.30</td>
<td>1.4</td>
<td>1.9</td>
<td>1.0</td>
<td>0.20</td>
<td>-</td>
<td>1.14</td>
</tr>
<tr>
<td>W.nr 1.2344</td>
<td>Hot work tools</td>
<td>0.40</td>
<td>1.0</td>
<td>0.4</td>
<td>5.2</td>
<td>-</td>
<td>1.35</td>
<td>1.0</td>
<td>1.98</td>
</tr>
<tr>
<td>W.nr 1.2363</td>
<td>Cold work tools</td>
<td>1.0</td>
<td>0.3</td>
<td>0.6</td>
<td>5.3</td>
<td>-</td>
<td>1.10</td>
<td>0.2</td>
<td>2.42</td>
</tr>
</tbody>
</table>

Table 3. Examples of tool steels. All elements in weight-%.

The W.nr 1.2738 steel is a good representative of the standard 300 HBW mould steel dedicated to plastic mould fabrication. The alloying contents in W.nr 1.2344 and W.nr 1.2363 make a low hydrogen weld method and high preheat temperatures necessary when welding. Repair welding of such steels is normally made in combination with direct tempering in furnaces after the welding operation is finished, followed by a slow cooling.

Again, the CE_{IIW} carbon equivalent is not a good reference when judging these steels’ weldability, as the empirical equivalent is based on a limitation of the amount of
martensite in the heat-affected-zone, and thereby a limitation of the hardness in mild steels. However, when welding Q&T-steels a soft (and large) heat affected zone will spoil the properties of the weldment. In contrary to weldments in the other steels described above, it is desired to minimize the loss in properties in the heat affected zone i.e to maintain a martensitic microstructure or a fine-grained acicular microstructure, depending on the application.

5. Experimental technique

In the present thesis hardenability of HSLA steels has been studied mainly using;

- welding simulation, resistive heating, of specimens 10x10x100 mm to enable production of Charpy-V specimens and to produce a synthetic HAZ for microstructural examinations.

- weld simulation using a Theta Dilatronic III dilatometer to study the phase transformation characteristics of the HAZ

- welding of real joints, to produce impact testing specimens, CTOD specimens and for microstructural examinations

- STEM (Scanning Transmission Electron Microscopy) studies of precipitation status of the grain coarsened HAZ using carbon replicas.

Synthetic HAZs were produced to facilitate the positioning of notches, microstructure evaluation specimens at a proper position to certify these being evaluated in a material volume which has undergone the desired time-temperature cycle. The temperatures were measured on the specimen surface using a spot-welded thermocouple. When performing these weld simulations specimens deviating more than 15°C from the aim temperature were scrapped. Also real HAZs were examined. However, there were difficulties to position the notches in a microstructure desired. When performing real weld tests, single bevel joints or K-type joints are preferred to make the notch positioning easier.

The weld metal properties were studied in real weldments. These were welded using MIG/MAG- and plasma-welding, whereafter impact testing specimens were taken to study the Charpy-V toughness. Furthermore, specimens for microstructural examination and for PASEM (Particle Analyzing Scanning Electron Microscope) studies of weld metal inclusions were taken adjacent to the impact testing specimens.
6. Microstructure and properties in the heat affected zone

6.1 General

The heat affected zone, HAZ, is of great concern in a weldment since it can contain a large variety of microstructures, geometries, stresses and thereby have different properties depending on where in the HAZ a test is made. The time-temperature cycle during the weld process, in combination with the chemical composition in the parent plate and its thickness governs the HAZ properties. When judging the weldability of a steel it is of great importance to consider the peak temperature reached in the HAZ which has a large influence on;

- the dissolution of microalloying precipitates in the grain-coarsened HAZ
- the austenite grain growth
- the cooling rate from the peak temperature.

It is, indeed, possible to look at the fully austenitized HAZ as a quencher in miniature.

When welding a structural steel, it is the delivery condition that governs the heat input used. Steels produced via normal rolling/normalizing/TMCP have to be welded using a high enough heat input to avoid too much martensite in the HAZ, while at the same time choosing a heat input that is low enough to avoid excessive austenite grain growth in the grain coarsened HAZ. These limits are given by the actual chemical composition of the steel. On the contrary, when welding a quenched and tempered steel a maximum heat input has to be chosen, thereby limiting the softening in the HAZ.

6.2 Austenite grain growth control

Control of grain growth in the HAZ is of great importance since the austenite grain size influences the hardenability and also the mechanical properties. As a good toughness is needed in most HAZs a small grain size is desired. The austenite grain growth in the HAZ is governed by the peak temperature, cooling rate and also by grain growth inhibitors as microalloy precipitates stable enough to survive the thermal cycle. To reduce the tendency for grain coarsening, microalloying nitrides or carbonitrides can be introduced into the steel matrix, but unfortunately very few of these are effective at the high temperatures which occur during welding. Titanium nitride, TiN, is the most common precipitate used which can be effective even at such high temperatures as are reached in the grain coarsened HAZ. Other possible particles reported to be stable at high temperatures are TiO and Ca(O,S) (10 - 11). However, when alloying with titanium a very small amount of Ti has to be used as the goal is to achieve small (~100 Å) TiN precipitates. With too high a titanium addition the nitrides start to precipitate at higher temperature and then become too coarse to control the austenite grain growth.

Gladman (12), has shown that equating the forces of austenite grain growth and the pinning force on austenite grain boundary movement exerted by precipitates in steel
matrix gives the following expression

\[ R_0 = r_c(1.5-2/Z)\pi/6f \]  

(Eq 8)

where \( r_c \) is the precipitate size

\( f \) is the fraction of precipitates

\( R_0 \) is the average austenite grain size

\( Z \) is the radii ratio between growing grains to the average grains

In this thesis work austenite grain growth was studied from two aspects, first as isothermal grain growth, using simulation techniques, for a given soaking time at a given temperature and secondly as grain size after different thermal cycles representing different parts of the HAZ, see Paper A and Paper B.

Investigations were carried out to evaluate the microalloy precipitates in some HSLA steels microalloyed with titanium, vanadium and niobium in various contents. In the as normalized condition these steels have precipitates smaller than 25 nm. Table 4 gives the chemical compositions of the steels evaluated.

<table>
<thead>
<tr>
<th>Steel</th>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>P</th>
<th>S</th>
<th>Cr</th>
<th>Ni</th>
<th>Cu</th>
<th>Mo</th>
<th>Al</th>
<th>Nb</th>
<th>V</th>
<th>Ti</th>
<th>N</th>
</tr>
</thead>
<tbody>
<tr>
<td>CMn</td>
<td>0.10</td>
<td>0.32</td>
<td>1.34</td>
<td>0.012</td>
<td>0.002</td>
<td>0.02</td>
<td>0.02</td>
<td>0.02</td>
<td>0.02</td>
<td>0.036</td>
<td>-</td>
<td>0.014</td>
<td>0.002</td>
<td>0.005</td>
</tr>
<tr>
<td>Ti</td>
<td>-</td>
<td>0.30</td>
<td>1.33</td>
<td>0.011</td>
<td>0.005</td>
<td>0.01</td>
<td>-</td>
<td>-</td>
<td>0.03</td>
<td>-</td>
<td>-</td>
<td>0.009</td>
<td>0.008</td>
<td>0.007</td>
</tr>
<tr>
<td>TiNb</td>
<td>0.11</td>
<td>0.29</td>
<td>1.31</td>
<td>0.009</td>
<td>0.007</td>
<td>0.02</td>
<td>-</td>
<td>-</td>
<td>0.004</td>
<td>0.023</td>
<td>0.021</td>
<td>0.013</td>
<td>0.007</td>
<td>0.005</td>
</tr>
<tr>
<td>TiV</td>
<td>-</td>
<td>0.33</td>
<td>1.40</td>
<td>0.011</td>
<td>0.008</td>
<td>-</td>
<td>-</td>
<td>0.03</td>
<td>0.04</td>
<td>0.006</td>
<td>0.052</td>
<td>-</td>
<td>0.055</td>
<td>0.009</td>
</tr>
<tr>
<td>TiNbV</td>
<td>0.10</td>
<td>0.30</td>
<td>1.38</td>
<td>0.010</td>
<td>0.006</td>
<td>-</td>
<td>-</td>
<td>0.02</td>
<td>-</td>
<td>0.032</td>
<td>0.023</td>
<td>0.054</td>
<td>0.008</td>
<td>-</td>
</tr>
<tr>
<td>LCTiNb</td>
<td>0.09</td>
<td>0.38</td>
<td>1.32</td>
<td>0.007</td>
<td>0.002</td>
<td>0.05</td>
<td>0.11</td>
<td>0.20</td>
<td>0.011</td>
<td>0.037</td>
<td>0.026</td>
<td>0.016</td>
<td>0.005</td>
<td>0.007</td>
</tr>
</tbody>
</table>

Table 4. Chemical compositions, all elements given in weight-%.

A calculation of the amount of element dissolved/precipitated can be obtained using the following equations.

\[ \frac{(w_{Me}^{st} - w_{Me}^{d})}{(w_{N}^{st} - w_{N}^{d})} = \frac{M(\text{Me})}{M(N)} \]  

(Eq 9)

\[ \log K(\text{MeN}) = A - B/T \]  

(Eq10)

where

\( w_{Me}^{st} \) is the amount of microalloy element in the steel

\( w_{Me}^{d} \) is the amount of microalloy element dissolved at the temperature \( T \)

\( w_{N}^{st} \) is the amount of nitrogen or carbon in the steel

\( w_{N}^{d} \) is the amount of nitrogen or carbon dissolved at the temperature \( T \)

\( M(\text{Me}) \) is the molar weight of the microalloy

\( M(N) \) is the molar weight of nitrogen or carbon

\( A \) and \( B \) are constants, and

\( T \) is the absolute temperature
This shows that at a peak temperature of $1350^\circ\text{C}$ only titanium nitride can be expected to be found in the HAZ. As other microalloying elements such as niobium and vanadium are added to the steel, the stability of titanium nitride at peak temperatures, as given above, diminishes. This is also demonstrated by an increase in austenite grain size in the grain coarsened HAZ as compared to the plain titanium microalloyed steel, see Table 5.

<table>
<thead>
<tr>
<th>Grade</th>
<th>Austenite grain size (µm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>CMn</td>
<td>255</td>
</tr>
<tr>
<td>Ti</td>
<td>50</td>
</tr>
<tr>
<td>TiNb</td>
<td>72</td>
</tr>
<tr>
<td>TiV</td>
<td>100</td>
</tr>
<tr>
<td>TiNbV</td>
<td>77</td>
</tr>
<tr>
<td>Low carbon TiNb</td>
<td>102</td>
</tr>
</tbody>
</table>

**Table 5. Austenite grain sizes in the grain coarsened weld simulated HAZs.**

$T_p = 1350^\circ\text{C}$.

As is demonstrated in Table 5, an extensive grain growth takes place in the plain CMn-steel. A vanadium addition to a Ti-microalloyed steel seems to have a larger influence on the precipitate stability than is the case when alloying with niobium. However, during cooling from the peak temperature vanadium re-precipitates at titanium nitrides more rapidly than niobium does. The results presented in Paper A show that after heating up to $1175^\circ\text{C}$, immediately followed by water quenching, more small particles were present in the TiNb-, and TiNbV-steels than in the TiV-steel. At $T_p=1350^\circ\text{C}$, however, all steels exhibited very few and small precipitates. This higher amount of small precipitates at $1175^\circ\text{C}$ in the TiNb- and TiNbV-steels may be an explanation for the smaller grain sizes in these two steels as compared with the TiV-steel.

Also the heating rate to the peak temperature in the HAZ influences the stability of the precipitates, which is demonstrated in Table 6. A clear difference in austenite grain size between the two TiNb-microalloyed steels is shown in Table 5 and Table 6. In this investigation the LC TiNb-steel had a titanium addition sufficient to balance the free nitrogen and not aimed at controlling austenite grain growth, which was the goal in the higher carbon TiNb-steel.

<table>
<thead>
<tr>
<th>Grade</th>
<th>$d\gamma$ (µm) at $dT/dt$ 0.5°C/s</th>
<th>$d\gamma$ (µm) at $dT/dt$ 3°C/s</th>
<th>$d\gamma$ (µm) at $dT/dt$ 30°C/s</th>
<th>$d\gamma$ (µm) at $dT/dt$ 300°C/s</th>
</tr>
</thead>
<tbody>
<tr>
<td>CMn</td>
<td>393</td>
<td>379</td>
<td>321</td>
<td>254</td>
</tr>
<tr>
<td>Ti</td>
<td>265</td>
<td>113</td>
<td>65</td>
<td>48</td>
</tr>
<tr>
<td>TiNb</td>
<td>226</td>
<td>143</td>
<td>76</td>
<td>70</td>
</tr>
<tr>
<td>TiV</td>
<td>318</td>
<td>79</td>
<td>91</td>
<td>103</td>
</tr>
<tr>
<td>TiNbV</td>
<td>188</td>
<td>148</td>
<td>72</td>
<td>80</td>
</tr>
<tr>
<td>Low carbon TiNb</td>
<td>353</td>
<td>265</td>
<td>115</td>
<td>103</td>
</tr>
</tbody>
</table>

**Table 6. Influence of heating rate on austenite grain growth.**
As pointed out earlier introducing Ca(O,S) and/or TiO in a steel matrix has been reported to be beneficial for reducing austenite grain growth in the grain coarsened HAZ (10 – 11), especially at higher temperatures. An investigation carried out, Paper C, shows that in the trials where the heating rates were varied, no major difference in austenite grain growth can be observed between the steels having low and high calcium contents if the heating rate is larger than 3°C/s. When tested using a higher peak temperature, 1425°C, a much higher heating rate, about 100°C/s, is required to minimize the differences in grain growth behaviour between the two steels examined. When welding, however, the differences in grain growth behaviour observed have no major importance since the heating rates in welding are typically within the interval 200 – 500°C/s. Figure 2 shows that the heating rate obtained in the specimens after weld simulation (175°C/s), which is slightly lower than in common welding, gives no significant difference in austenite grain size between the two steels studied.

![Figure 2. Austenite grain growth in Ca-added steel, Paper C.](image)

6.3 Microstructures in HAZs of mild steels

Control of the HAZ microstructure is necessary due to its influence on the HAZ properties. When welding, brittle microstructures like lath martensite and bainite should be avoided in order to reduce e.g. the cold cracking risk. The microstructure obtained in the HAZ depends on a complicated interaction between steel chemistry, austenite grain size, precipitate status, and cooling rate from the peak temperature reached.
The cooling rate is determined by the heat input, groove configuration, plate thickness and ambient temperature. In mild steels a low cooling rate is beneficial to promote polygonal ferrite through a low super-cooling of the HAZ. Low cooling rate may also influence the austenite grain size from which the austenite is decomposed. A low rate will then result in an increased grain size which in turn normally increases the amounts of intra-granular ferrite and bainite on the expense of polygonal ferrite due to fewer nucleation sites for the latter. A high super-cooling promotes intra-granular ferrite, bainite or martensite. It is obvious that control of austenite grain size is important, especially when welding with high heat inputs. To obtain a small austenite grain size in order to promote polygonal ferrite instead of bainite. The principles and also practical control of grain growth was discussed in § 6.2. The chemical composition of the steel matrix will also affect the microstructure through its influence on the start temperature of the decomposition of austenite. If this temperature is lowered the amount of polygonal ferrite will be diminished, or even suppressed, such that acicular ferrite, bainite or lath martensite are promoted.

When evaluating the weldability of a steel it is of great importance to have knowledge of its CCT diagram (Continuous Cooling Transformation), normally constructed using a peak simulation temperature of 1350°C. An extensive study, Paper D, was carried out to establish the influence of microalloying of plain CMn-steels, and how these steels respond on different weld thermal cycles. The chemical compositions of these steels are suitable for producing heavy plates in S355 (355 MPa constructional steel) having thicknesses of 10 – 15 mm. The results are summarized in Figure 3, in which the hardenability has been defined as the cooling rate, $\Delta t_{8/5}$, required to give 50% martensite in the microstructure, i.e a link towards the Grossmann formalism.

These results clearly demonstrate the positive effect titanium nitride on grain growth inhibition which gives a reduction of the hardenability, especially at the highest carbon contents studied.
Figure 3. Cooling time, $\Delta t_{8/5}$, required to achieve 50% martensite in microstructure. Paper D.

The lowest heat inputs used in practical welding commonly give $\Delta t_{8/5} \approx 8-10$ s, which shows that most of the steels investigated can be welded without preheat in combined thicknesses up to about 40 – 50 mm. Furthermore, a smaller austenite grain size may be expected in real HAZs than when weld simulating in a dilatometer. This can be interpreted as due to the temperature gradient in the real welds towards the base material which gives a continuous decrease in the grain size network. It is probable that the adjacent finer grain structures can to some extent restrict the grain growth in the high temperature zone next to the fusion line. Since no temperature gradients, or at least only very small ones, are obtained within the weld simulation specimens, because of their small size and the heating method used (induction), unrestricted grain growth occurs and a coarser final austenite grain size is obtained, which will give a higher hardenability as compared with a real weld HAZ.

As was pointed out in Paper D most of the carbon equivalents found in the literature do not incorporate niobium when considering the hardenability. Only Yurioka (5) takes niobium into account. In many cases the equations proposed by (5)
seem to overestimate the HAZ-hardness, especially when titanium is introduced into
the steel matrix to control the austenite grain growth during the weld thermal cycle.
Due to the fact that niobium and titanium are commonly not taken into account when
judging a steels weldability a formula was presented pertaining to describe the
hardness in HAZ of micro-alloyed HSLA-steels having low levels of residual
elements, as is the case in modern steelmaking. The prediction was established for
cooling rates within the interval $\Delta t_{8/5} = 10-300$ s, since these cooling rates include
manual metal arc welding (MMA) where $\Delta t_{8/5}$ is about 10 s, and the longest cooling
times $\Delta t_{8/5}$ represent high energy welding. As pointed out earlier, see Figure 3, faster
cooling rates than $\Delta t_{8/5}$ about 10 s are normally not of interest since these result in too
high a HAZ-hardness and too much brittle microstructural components in the HAZ.

$$HV_5 = 487C+113Mn+44(Cr+Ni+Cu)+106V+606Nb-7.28Ti/N+876/\Delta t_{8/5}-46$$

The proposed formula predicts the HAZ-hardness with an accuracy of ± 15 HV5. The
relationship is valid for steels having carbon contents between 0.08 and 0.18%,
silicon contents around 0.3%, manganese contents around 1.4 – 1.5% and microalloy
contents as normally used.

6.4 Mechanical properties of mild steel HAZs

As the carbon equivalents for judging a steels weldability give no prediction of what
mechanical properties can be expected in the HAZ, tests have to be carried out to
determine the necessary properties. Of great importance is the toughness in the HAZ
as this may have an austenite grain size which, normally, is much larger than in the
parent plate. Steel development has been focused on minimizing the grain growth in
the HAZ when subjected to welding, mainly via a microalloy addition of titanium to
form titanium nitride, TiN. As the titanium nitride precipitates in austenite it does not
give any discernible precipitation strengthening, such as is achieved from niobium
and vanadium which normally are precipitated in ferrite. No drawbacks on toughness
from TiN precipitates are expected. A positive influence on toughness from TiN is
achieved as it minimizes the austenite grain growth.

In Paper B, C, E and F the control of toughness in HAZ is discussed. The most
effective parameters to control are;

- the austenite grain size which has a direct influence on toughness

- the amount of free nitrogen in the HAZ.

Impact toughness shows a linear relationship with the inverse square root of the
austenite grain size i.e the smaller austenite grain size the higher the toughness.
Furthermore, depending on the cooling rate from the peak temperature, smaller
austenite grain sizes promote transformation into polygonal ferrite and/or acicular
ferrite which are beneficial to toughness. As pointed out earlier, additions of niobium
and/or vanadium to a titanium-microalloyed steel decrease the precipitate stabilities,
thereby increasing the HAZ austenite grain size, which is harmful to toughness. The idea of utilizing Ca(O,S) as a grain size inhibitor showed no major influence of the austenite grain size when using high heating rates, although the introduction of Ca(O,S) turned out to increase the $Ar_3$-temperature and thereby a reduction of martensite was observed in the microstructure. Furthermore, the lamellar microstructures turned out to be finer when Ca(O,S) particles were introduced.

In Paper B a linear regression was made of ITT$_{27J}$ versus austenite grain size and the amount of polygonal ferrite (PF) for the steels examined. The equation is as follows:

$$\text{ITT}_27J = -67 - 0.28\text{PF} + 66d^{0.5} \text{°C}$$  \hspace{1cm} (Eq 11)

where the PF content is in percent and the grain size is in $\mu$m. Figure 4 shows the correlation between calculated and measured ITT$_{27J}$ for the steels investigated. Here we have a first example on a Petch type relationship. In Paper J a to some extent similar relationship of toughness versus grain size of the rolled and quenched steel base metal was obtained. Once again this demonstrates that the HAZ is actually a quenched base material as to structure.

![Figure 4. Correlation between calculated and observed ITT$_{27J}$.](image)

Free nitrogen in the HAZ is extremely harmful to the toughness. In Paper F, an increase of the impact transition temperature at the 27J energy level (ITT$_{27J}$) by 2-5°C/10 ppm free nitrogen was reported, which is in agreement with data reported
in the literature (13). The nitrogen content in a steel should be kept as low as possible and preferably be stabilized utilizing titanium or an other strong nitride former.

Another way to control HAZ toughness is to adopt, if possible, a multi-run welding schedule where the heat input from a subsequent pass normalizes the earlier one. In Paper E, it was demonstrated that when adopting such a welding technique, in combination with vanadium-microalloying where the vanadium nitride precipitates have the function to tie up free nitrogen, very low ITTs, -80°C to -100, were then observed due to a very fine grained ferritic microstructure.

A fitness for purpose study for a TiV structural steel now used for bridges, ships and the like has been performed in an addendum to Paper B. Critical crack size as based on CTOD on one side and on Charpy-V on the other is compared.

7. The weld metal deposit

7.1 General

When considering the weld metal properties in mild steels and in medium strength steels there are generally no problems to find a suitable filler material to fulfil the strength requirements. However, when welding high strength steels, yield strength of more than 960 MPa, hardly any filler materials are available when matching weld metal strength is desired. The lack of suitable filler materials is mainly due to a drastic increase of the cold crack risk as high alloy contents are normally used when producing high strength weld metal deposits.

Within the thesis project, solid filler wire was produced using the chemical composition of one of the developed tool steels. When welding using this filler material a matching weld deposit can be produced, not being extensively sensitive to cold cracking as a totally different alloying approach has been made.

7.2 Control of weld metal properties

To control weld metal properties care has to be taken of a weld deposit concerning:

- inclusion content
- microstructure
- alloy content
- residual content

The oxygen content is of particular interest in a weld metal deposit since it can have an influence on the microstructure, and it also influences the upper shelf energy in toughness tests via the inclusion sizes and distributions. The alloy content also is of great importance as it influences the weld metal microstructure and contributes to its tensile properties. Manganese has a two-fold influence on the properties since it has an influence on the $A_3$-temperature (affects the amount of acicular ferrite) and also
contributes linearly to the yield strength via solid solution hardening. If too high a manganese content is added to the weld metal, bainitic microstructures start to appear in the steel, which results in a lower toughness. The manganese content in weld metal can be controlled via the degree of interfusion combined with the choice of filler material manganese content and the manganese losses in the arc. With MAG-welding, the shielding gas can also be used to control manganese loss via the choice of gas mixture. Grong and Christensen (14) have shown that the losses of both silicon and manganese increase when increasing the CO₂ content in the shielding gas. Hannerz (15) has shown in an earlier part of the SEGA project that in SAW weldments a minimum in impact transition temperature of the weld metal is observed within a range of manganese content of 1.0 to 1.6 %. For lower weld metal carbon contents the minimum is shifted towards the upper manganese content (1.6 %). In the same study, Hannerz also reported a low Si/Mn-ratio to be beneficial for toughness. Numerous authors have proposed that the inclusion content in weld metal might affect the microstructure. Abson et al (16) pointed out that if the right sizes of inclusions can be achieved, these can act as nucleation sites for acicular ferrite, thereby increasing the weld metal toughness. In Paper G and H investigations were carried out of the influence of varying manganese and silicon contents on weld metal properties when using two different weld methods were carried out. No influence of the inclusion sizes on the amount of acicular ferrite could be detected see Figure 5.

![Figure 5. Acicular content versus area fraction of inclusions smaller than 0.1 µm.](image)

However, as expected, an influence of the inclusion content on the upper shelf energy was shown, see Figure 6.
$10 \log N_a(D) = kD + l$  
(Eq 12)

Where:
$N_a(D)$ is the number of inclusions within a given size range
$D$ is the actual inclusion size
$k$ and $l$ are constants

In the studied MAG-weldments the upper shelf energy in Charpy-V-testing was also shown to vary according to equation 13, see also Figure 7;

$E_{us} = 10.5 - 60.3Si - 14.9Mn - 0.3Mn/Si - 792O - 15.8Q + 2.3AF + 2.5PF + 1.8AC$ (J)  
(Eq 13)

where alloy contents in weight-%
heat input ($Q$) in kJ/mm
and AF = acicular ferrite, PF = polygonal ferrite, AC = bainite/widmanstätten ferrite
given in %. The relationship is valid for weldments in mild steels, using mild steel
ciller metal and heat inputs in the range 10 to 25 kJ/cm.
The impact transition temperature at 27 J in MAG-weldments, ITT$_{27}$, is influenced by the manganese content as shown in Figure 8. The optimum around 1.1 % manganese depends on the amount of acicular ferrite in the microstructure. At lower manganese contents polygonal ferrite dominates, while at higher manganese levels a bainitic microstructure is dominating.
8. Summary of the first part

The first part of the thesis deals with the hardenability, and mechanical properties, of the HAZ in low alloyed and in micro-alloyed steels. Furthermore, the first part also discuss the microstructure and toughness of weld metal in mild steel weldments.

In Paper D, the influence of microalloying on the hardenability in HSLA steels is evaluated and shows the possibilities of controlling the hardenability by using proper combinations of titanium and nitrogen pertaining to control of the austenite grain size which thereby counteracts the negative influence of other microalloying elements on the hardenability.

Paper A, B, C, E and F discuss in more detail how to control structures and properties, primarily using titanium nitrides to control the austenite grain size during a thermal cycle. Furthermore, the possibility of using Ca(O,S) to control grain size at high peak temperatures (1425°C) is evaluated. The grain size control mechanism has been applied to mild and low alloyed steels to increase their mechanical properties in the grain coarsened HAZ after welding, and the papers show that very attractive results can be achieved.

Also the possibilities of controlling weld metal microstructure in mild steel weldments have been studied after MAG- and plasma-welding. It is clearly pointed out that acicular ferrite is preferred in the weld metal microstructure to optimize the weld metal properties, especially the toughness.

The results from the findings in Paper A to F were used by the present author and colleagues in the 1990’s to improve the abrasive resistant steels, HARDOX, (and armouring plate) as to weldability considering also earlier production knowledge from Oxelösund steel works as outlined in Paper I where the welding of abrasive resistant steels is discussed and welding procedures are recommended.

A fitness for purpose study based on the same principles as with the tool steel was made on the most commonly used TiV structural steel with the use of both CTOD and Charpy-V results in the addendum to Paper B, with the same principles as used for TOOLOX in Paper K.

Finally, as outlined in Paper I steels with very much improved steel cleanliness and still higher carbon equivalents were developed using very much the same principles as given above, These steels are tool steels intended for plastic moulds having a mirror polished surface, or photo-etched surface, but still weldable which is of great importance since such steels are often repair welded. This will be discussed in the next chapter.
9. The development of two new tool steels

9.1 General

Moulds for plastic fabrication are traditionally produced using quenched and tempered tool steels having high alloy contents, such as the P20/W.nr 1.2311/1.2738 or H13/W.nr 1.2344, which are developed for oil quenching. The two first grades are delivered as quenched and tempered with a nominal hardness of 300 HBW. However, when producing moulds in higher hardmesses, above approximately 38 HRC, the traditional mould material, H13, is normally supplied in a soft-annealed condition. After machining, a tool made in this steel is quenched and tempered to the hardness level desired, normally within the hardness range 38 – 55 HRC. However, during the heat treatment operation shape changes often take place due to;

- ferrite → austenite transformation during heating to the austenitizing temperature desired.
- austenite → martensite transformation during the quenching, and
- the decomposition of martensite and precipitation reactions which take place when tempering the steel.
- non-uniform temperature leading to thermal stresses and plasticity.

These changes depend on mould steel chemical composition, austenitizing temperature, cooling rate during the quenching operation, mould geometry etc. After receiving the mould from heat treatment the mould maker has to adjust the shape of it into its final dimensions. Furthermore, the cracking risk, which may be quite high due to the high carbon and alloy contents, which shall not be neglected.

The machining technique has seen a rapid development during the last years which has given equipment suitable to machine also hard materials. This possibility has led to the development of a new generation of tool steels, brand names TOOLOX 33 and TOOLOX 44 respectively, introduced on the market mid 2002. The main objective when designing these grades was to use the advanced metallurgy/rolling and heat treatment facilities at SSAB Oxelösund to produce mould steels delivered in as quenched and tempered condition having two hardness levels 300 HBW (TOOLOX 33) and 450 HBW (TOOLOX 44) respectively. As these grades need no further heat treatment, a mould is ready to set into production directly after machining or after being surface treated if necessary. Of these steels, TOOLOX 44 represents an entirely new hardness level of pre-hardened tool steels on the market. The shorter lead times possible in mould manufacturing make these steels interesting in technical as well as in economical terms. The new grades possess an excellent machinability, in combination with a far better weldability, as compared to the tool steels commonly produced for the same purpose.

The new steels developed, as will be described in this paragraph, have been patented in Sweden (Paper L).
9.2 Demands put on the new steels

In 2000, when the development of the new steels started, a number of important demands put on tool steels, which have to be fulfilled in these grades, were listed and compared to the demands put on the regular quenched and tempered HARDOX and WELDOX grades produced by SSAB Öxelösund AB, see Table 7.

<table>
<thead>
<tr>
<th>Property</th>
<th>HARDOX / WELDOX</th>
<th>Tool steels</th>
</tr>
</thead>
<tbody>
<tr>
<td>Weldability</td>
<td>Extra high</td>
<td>Low</td>
</tr>
<tr>
<td>Bendability</td>
<td>&quot;&quot;</td>
<td>&quot;&quot;</td>
</tr>
<tr>
<td>Machinability</td>
<td>Low to moderate</td>
<td>Extra high</td>
</tr>
<tr>
<td>None</td>
<td>&quot;&quot;</td>
<td>&quot;&quot;</td>
</tr>
<tr>
<td>Shape stability when machining</td>
<td>None</td>
<td>-&quot;&quot;</td>
</tr>
<tr>
<td>Material cleanliness</td>
<td>Moderate to high</td>
<td>-&quot;&quot;</td>
</tr>
<tr>
<td>Toughness</td>
<td>High</td>
<td>-&quot;&quot;</td>
</tr>
<tr>
<td>Segregation control</td>
<td>Moderate to high</td>
<td>Extra high</td>
</tr>
<tr>
<td>None</td>
<td>&quot;&quot;</td>
<td>&quot;&quot;</td>
</tr>
<tr>
<td>Polishability</td>
<td>None</td>
<td>-&quot;&quot;</td>
</tr>
<tr>
<td>Photo-etching ability</td>
<td>&quot;&quot;</td>
<td>-&quot;&quot;</td>
</tr>
</tbody>
</table>

Table 7. Demands put on different grades.

A survey was also performed to evaluate the two most common tool steels at the market, the P20 and the H13 grades. These steels have high alloy contents, see Table 8, and are designed for oil quenching.

<table>
<thead>
<tr>
<th>Grade</th>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>P</th>
<th>S</th>
<th>Cr</th>
<th>Mo</th>
<th>V</th>
</tr>
</thead>
<tbody>
<tr>
<td>P20</td>
<td>0,40</td>
<td>0,30</td>
<td>1,45</td>
<td>max</td>
<td>0,035</td>
<td>1,95</td>
<td>0,20</td>
<td>-</td>
</tr>
<tr>
<td>H13</td>
<td>0,40</td>
<td>1,0</td>
<td>0,4</td>
<td>max</td>
<td>0,030</td>
<td>5,2</td>
<td>1,35</td>
<td>1,0</td>
</tr>
</tbody>
</table>

Table 8. Typical chemical compositions, elements in weight-%.

P20 is delivered as quenched and tempered having a hardness within the interval 280 – 325 HBW, while H13 is delivered in as soft-annealed condition. Depending on the application H13 is heat treated to hardnesses between 38 - 55 HRC. A major drawback of these steels is that they have a poor toughness as compared to the toughness in the WELDOX/HARDOX grades. P20, which has a typical toughness of 15 – 20 J at room temperature can be compared to the toughness in WELDOX 700 (which has a comparable hardness) of about 200 J at room temperature. H13 has a typical toughness of 8 – 15 J at room temperature, when heat treated to approx.
42 HRC, which shall be compared to a typical toughness of 70 – 90 J in HARDOX 400. This information made it obvious that when designing the new tool steel grades, a choice of chemical composition and process route in the production can give properties which outrun the ones of P20/H13 with respect to plastic mould applications. Furthermore, the P20 is often used in machine applications when high strength is required and a new grade having much higher toughness than P20 was then a goal.

Also the different production steps, necessary to produce the new tool steel grades at the mill were scrutinized giving the following demands:

- Ability to produce extremely clean steels and to continuous cast them with as low segregation levels as possible.

- Effective rolling, also in plate thicknesses up to 130 mm without any porosities.

- A well functioning heat treatment facility (quencher and tempering furnaces).

The conclusion was that all three production steps should be able to fulfil the demands put.

As is shown in Table 7 the major demands against which HARDOX/WELDOX are developed to fulfil are weldability and bendability. These properties can be achieved utilizing low alloy contents, and a steel cleanliness high enough to facilitate good bendability properties. When the development of the new steels began, the steel production was well prepared to meet the requirements to produce ultra-clean steels, and also to produce slabs having a very good segregation control thanks to the possibility to use soft reduction in the continuous casting process. Furthermore, the development of steels having an excellent weldability had also forced the steel production to produce steels within very narrow tolerances in chemical composition, which simplified the tool steel development. It may look a little bit confusing that the weldability is not a top demand in the tool steels developed, but this has its origin in when producing moulds and tools the manufacturers normally do not bother about the weldability. Usually, welding is in these cases different kinds of repair welding of quite small units, as compared to the structures welded in HARDOX/WELDOX. The mould manufacturers are used to weld in the traditional tool steels, having very high alloy contents, where preheat temperatures of 325 – 375°C might be necessary, followed by a controlled slow cooling. In Table 9, the main demands on steel design and production to meet the requirements are listed.
Table 9. Methods to fulfil the demands put on the steels developed.

In the final design of tool steels the main questions to clarify were;

- austenite grain size control to optimize toughness
- hardenability, to enhance production of heavy sizes
- material cleanliness and segregation control, to permit polishing to mirror surface quality and also to make photo-etching possible
- material cleanliness to maximize the fatigue properties
- mechanical properties, i.e optimization of hardness and toughness
- low alloy content to enhance weld repair of moulds, which gives an advantage against the traditional tool steels
- optimization of the heat conductivity by using as low alloy contents as possible to minimize plastic component production cycle times, as these are dependent on heat transfer in the mould.

The goal was to produce two plastic mould grades;

- one having a target hardness of 310 HBW, named TOOLOX 33, to be launched against P20/W.nr 1.2311/1.2312
- the second one with the target hardness 450 HBW (45 HRC), named TOOLOX 44, which is an entirely new steel introduced on the market.
- Both grades shall be delivered in as quenched and tempered condition when delivered from the mill, and not intended to be further heat treated. This makes it necessary to produce these grades that they have an excellent shape stability when machined.
9.3 Steel and plate production

The chemical compositions of the two grades were determined using the experience from the development of HARDOX/WELDOX. To establish the chemical composition desired hardenability according to Grossmann (1) and to Just (6) was made, see Paper J. From the compositions determined slabs were produced. As compared to the traditional tool steels, see Table 8, the new steels have low alloying contents, and these are given in Table 10.

<table>
<thead>
<tr>
<th>Grade</th>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>Cr</th>
<th>Mo</th>
<th>Ni</th>
<th>V</th>
<th>Ti</th>
<th>Ew</th>
</tr>
</thead>
<tbody>
<tr>
<td>TOOLOX 33</td>
<td>0.26</td>
<td>0.60</td>
<td>0.90</td>
<td>1.18</td>
<td>0.40</td>
<td>0.70</td>
<td>0.12</td>
<td>0.016</td>
<td>0.80</td>
</tr>
<tr>
<td>TOOLOX 44</td>
<td>0.30</td>
<td>0.60</td>
<td>0.90</td>
<td>1.23</td>
<td>0.80</td>
<td>0.70</td>
<td>0.14</td>
<td>0.016</td>
<td>0.93</td>
</tr>
</tbody>
</table>

Table 10. Typical chemical compositions and carbon equivalents.

Heavy plates were produced using conventional hot rolling, air cooling to room temperature followed by austenitizing at 925°C and water quenching. Thereafter, the proper tempering temperature was evaluated at specimens cut from plates, and the plates tempered at the temperature determined. The tempering response of the two grades are given in Figure 9.

Figure 9. Tempering response of the two steels produced.
The plate production is shown schematically in Figure 10.

![Figure 10. Plate production, schematic](image)

The work concerning austenite grain size control in the weld heat affected zone presented in Paper B, C and F was used to determine the Ti-microalloying needed to control austenite grain size of the tool steels developed during austenitization prior to quenching, as well as to control the austenite grain size during welding. This was made pertaining to control the toughness of the materials via a small austenite grain size. As pointed out in Paper J, the impact toughness of these tool steels developed show a linear relationship according to the inverse square root of the austenite grain size, see Figure 11. When producing heavy forged bars of the two grades the long soaking times at high temperatures, in combination with slow cooling after finished forging gives a coarse austenite grain size (about 250 – 500 µm) as compared to the grain sizes achieved after plate rolling (about 25 µm). To improve the impact toughness of the forged bars other precautions must be taken to control the austenite grain size, which will be studied in the future.
Figure 11. Relation between austenite grain size and impact toughness.

9.4 Machinability

Traditionally, steels aimed to machining are re-sulphurized to enhance their machinability. As pointed out in Paper J, the machinability of TOOLOX 33 is far better than the machinability of the regular, re-sulphurized (about 0.06%), W.nr 1.2312 despite the far lower sulphur level of TOOLOX 33 (typically 15 – 20 ppm), see Figure 12.
The results show that the opinion that sulphur influences the machinability must be restricted to when comparing two equal chemical compositions, as W.nr 1.2311 (low sulphur) and W.nr 1.2312 (re-sulphurized). A comparison between TOOLOX 33 and W.nr 1.2312 must be made according to the amount of carbides, cementite etc., in the microstructure. The mechanism behind these differences is not yet fully understood, but there are indications that the amount of carbides in matrix governs the wear of the cutting tool edges (17).

9.5 Welding of the new steels

9.5.1 Repair welding

Furthermore, it is also necessary to have the possibility to weld repair a mould when it is on maintenance, or if anything goes wrong during the mould production. As pointed out earlier the steel cleanliness and segregation level are important to control to permit the steel to be used in applications where high demands are put on the surface i.e. polishability and photo-etchability. These properties must also be met after a repair welding of a mould and the common practice when repair welding a mould is to preheat to the temperature prescribed, weld using a filler material having equal or nearby equal chemical composition, cool to room temperature and thereafter temper at a proper temperature with respect to the actual steel. The final test of whether a weld repair has succeeded is when the mould is re-polished and re-etched.
A test was carried out to check the properties after repair welding. Filler material, made from TOOLOX 44 having Ø 1.0 mm and 1.2 mm, was produced as solid wire with a chemical composition as given in Table 8. Test samples of TOOLOX 33 were repair welded using Micro-TIG, polished and etched using a fine grid which reveals any defects from mismatch between base material, HAZ and weld metal. As is shown in Figure 13, no imperfections neither in the HAZ nor in the weld deposit could be revealed.

![Figure 13. Repair welded and photo-etched TOOLOX 33.](image)

9.5.2 Welding of TOOLOX 33

Test welding using TIG was performed on TOOLOX 33 show excellent toughness results as shown in Paper K., see Figure 14 and 15. When test welding TIG-welding was chosen as when welding in tools steels this is the commonly used method.
A failure assessment of the base material according to BS 7910 was carried out on TOOLOX 33 using the data from plate production of TOOLOX 33 which is 109 J at +20°C, see Paper K. As a comparison an assessment was carried out on the traditional tool steel W.nr 1.2311 which is comparable to TOOLOX 33. The W.nr 1.2311 has a typical toughness of 15 J at room temperature. A comparison shows that the much larger tolerance to defects of TOOLOX 33 due to its higher toughness. When looking at a weld which has not been stress relieved, the same assessment of TOOLOX 33 shows a reduction of critical defect size from 17.3 mm to
3.1 mm due to the inherent residual stresses from the welding operation when assuming a semi-circular flaw of \(a/2c = 0.5\)

In future studies of the properties of a stress relieved weldment is important to carry out pertaining to establish the max. defect size permitted.

Today the results clearly demonstrate that the new TOOLOX 33 can be repair welded using the TOOLOX 44 chemical composition in a filler wire. The ultra-clean wire made from slabs dedicated to production of TOOLOX 44 yields a weld metal where the prior austenite grain boundaries are much cleaner as compared to grain boundaries in a weld deposit produced by regular filler wires. The higher cleanliness lowers the risk of decreased grain boundary cohesion strength and thereby decrease the risk of inter-granular fracture. Another benefit from such a clean filler wire is the much lower nitrogen and oxygen contents as compared to regular wires. The present filler wire has a nitrogen content of 36 ppm. The oxygen content has not been measured but the actual steelmaking process route of equal grades yields oxygen contents below 10 ppm. In combination with the TIG process which gives a very low oxygen increase in the weldments low amounts of inclusions which may disturb the toughness, as discussed in Paper G and H, this new approach of filler wire chemical composition gives the attractive properties achieved.

In the addendum to Paper B a similar fitness for purpose study was made for a structural TiV micro-alloyed steel using both CTOD and Charpy-V. Naturally much larger critical flaws can be tolerated in the structural grade than in a quenched and tempered tool steel.

9.6 Concluding remarks on steel development

The goals set for the development of the two new steels, as described in § 8.1, have been met, and details of the development work are presented in Paper J. As such good steel properties were achieved application for a Swedish patent was made and the Swedish Patent Authority has issued a Swedish patent, Pat.No 524133, for these new steels, see Paper L. Furthermore, another concept of alloy design on filler wire was revealed when evaluating the welding properties of TOOLOX 33.

10. Welding of super high strength steels

10.1 General

When welding steels having yield strengths above 900 MPa, no filler material is available at the market from the major manufacturers of consumables. One major drawback when developing filler materials for such grades is that, when using the common alloy philosophy, high alloying contents are used which result in severe cold cracking problems. Commonly, steels having such high yield strengths are welded using softer filler material. As filler material was produced during the development of the two new tool steels they were tested in welding of super high strength steels.
10.2 Welding of steels of Y.P 900 MPa

From the encouraging results from the test welding of TOOLOX 33 the new filler wire was tested in weldments of commercial super high strength steels, WELDOX 900 and WELDOX 1100 having minimum guaranteed yield strength of 900 MPa and 1100 MPa respectively.

Welding of WELDOX 900 using the TIG-method, as shown in Paper K, gives a weld metal that match the base metal properties, while the MAG-weldments of WELDOX 1100 do not match the base metal properties with regard to weld metal toughness. The main reason for this behaviour is that the MAG-process gives much higher oxygen contents in the weld metal, which results in a large amount of inclusions in the weld metal, as was also discussed in Paper G.

Due to productivity and economical reasons TIG-welding is only suitable when welding thinner gauges, but if high demands are put on the weld metal in a lightweight construction the filler wire produced from the chemical composition of TOOLOX 44 may be an attractive alternative.
11. REFERENCES


12. Summary of appended papers


This work was carried out in co-operation between the authors where Hansson made the weld simulations and also took part in the electron microscopy work.

Precipitation status in microalloyed steels has been characterized in the normalized condition as well as after various weld thermal cycles. The chemistry alterations of individual particles and compositional gradients within one and the same precipitate were measured in field emission gun scanning transmission electron microscope. Energy loss spectroscopy has been performed for the light element (carbon and nitrogen) analyses. Particle size, frequency, and distribution were established using carbon extraction replicas. The particles were classified in histograms by an image analyser.

It is shown that the presence of additional microalloying elements (Nb and/or V) in a Ti-microalloyed steel affects the persistence of TiN, basically of (Ti,Me)N-type, on rapid heating to increasing temperatures. It is clear that TiNbV favour the stability. When V is used as an alloying element the (Ti,V)N was found to have an inferior stability to TiN. The experiments revealed, however, a rapid reprecipitation of VN on TiN during cooling of the TiV-steel, but almost no reprecipitation of Nb(C,N) was detected in the Nb-bearing steels.


Paper B was carried out within a cooperation project between Sweden and China on weldability studies of titanium microalloyed HSLA steels, where the authors co-operated in experimental work, evaluation and report writing.

The results show that a small addition of Ti (0.01%) to CMn-steels based on commercial production can effectively control the austenite grain growth during welding. The presence of Nb and/or V in a Ti-microalloyed steel lessens the excellent grain coarsening inhibition derived from TiN only. Titanium addition decreases hardenability, while a Nb and/or V microalloying increases this. The effect of V addition on HAZ hardness is clear only for high heat input conditions. Niobium increases hardness over the whole heat input range from low to high inputs. Presence of Nb together with V further increases this effect.

The impact transition temperature, measured at an absorbed energy level of 27J (T_{27}) at the fusion line of the high heat input weldments increases with increasing austenite grain size, and decreasing amount of polygonal ferrite. The same behaviour was also revealed in the COD-tests.
A low cold cracking propensity was observed in all Ti-microalloyed steels tested using the Implant method. In fact, these steels had to be provoked to crack by welding with a high hydrogen acid electrode. The results indicate a lower cold crack tendency with decreasing austenite grain size and decreasing carbon content.


In Paper C the possibility to reduce the austenite grain size in the grain coarsened region of the HAZ via calcium additions to a Ti-Nb-V-microalloyed steel was elucidated. Ca(0,S)-inclusions, which are thermodynamically more stable than TiN should not be dissolved to the same extent during high energy welding. Such a reduction in HAZ austenite grain growth should be instrumental in achieving an increased toughness in this region, or alternatively in giving the possibility to use a higher welding energy without risk for unacceptable austenite grain growth in the HAZ.

In this paper, the austenite grain growth behaviour has been studied,
- through isothermal heat treatment in the temperature range 900-1300°C,
- via welding simulation with two peak temperatures, 1350°C and 1425°C, and
- as a function of the heating rate (0.5 – 250°C/s) to the peak temperature 1350°C and 1425°C followed by cooling of 10°C/s.

In this connection, two Ti-Nb-V-microalloyed steels were used, one of which was calcium treated (33 ppm Ca), while the other had not been subjected to any intentional calcium treatment. An impurity content of 13 ppm Ca was, however, measured for this latter steel. The results show that the low calcium steel shows a diminished austenite-grain-growth propensity compared with the high calcium steel during both welding simulation and the isothermal heat treatment. On the other hand, no difference in the heating rate effect on the austenite grain growth could be detected between the two steels.

The variations in the prior austenite grain size between the two steels is also reflected in a difference in the degree of refinement of the transformed constituents in the HAZ microstructure; however, the proportions of the individual constituents are almost the same for the two steels. The best impact toughness in weld-simulated material was found for the low-calcium steel and the toughness did not deteriorate when the higher peak temperature was used. The high-calcium steel, which was characterized by an inferior impact toughness, exhibited a considerable difference in the impact-transition-temperature depending on the peak temperature in the thermal cycle. The alterations in toughness are most probably related to the observed variations in the microstructure depending on various degrees of austenite grain growth. One possible explanation for the differences between the two steel variants is that the low calcium steel contains higher levels of titanium and nitrogen, which result in a greater amount of un-dissolved TiN at temperatures near the peak.

In Paper D a study has been carried out to investigate the HAZ hardenability of microalloyed steels with different carbon levels using experimental CCT diagrams. The study included plain CMn-steels as well as Nb-, NbTi-, VTi-, NbVTi- and NbV-microalloyed variants having carbon contents of 0.08%, 0.12% and 0.18%. The results show that Nb-microalloying gives rise to an increase in hardenability referred to the plain CMn-steel, which implies that Nb should be included as a hardenability-increasing element in the carbon equivalent. Microalloying with V gives a lower amount of martensite as compared with the CMn-steel for rapid cooling conditions but nevertheless results in a higher hardness. This hardening effect is evident in carbon equivalent expressions which include vanadium. Combination of Nb+V gives rise to a synergistic effect on hardenability which includes both martensite content and hardness. Ti-microalloying reduces hardenability as a result of the grain growth inhibiting effect of TiN particles during the welding cycle. This effect was especially evident for the TiV-steel with 0.08% C where the microstructure for \( \Delta t_{8/5} = 20 \text{s} \) consisted of 80% polygonal ferrite.

An equation for calculation of HAZ hardness for \( \Delta t_{8/5} \) between 10 sec and 300 sec has been determined as:

\[
HV_5 = -46+478 \cdot C+113 \cdot Mn+44 \cdot (Cr+Ni+Cu)+106 \cdot V+600 \cdot Nb-7.28 \cdot (Ti/N)+876/\Delta t_{8/5}
\]

The relationship is valid for steels having carbon contents between 0.08 and 0.18%, silicon contents around 0.3%, manganese contents around 1.4 – 1.5% and microalloy contents as normally used.


In Paper E gives an investigation which was carried out to elucidate the influence of varying heat input on the toughness of micro-alloyed steels with different nitrogen Levels. The behaviour of V-micro-alloyed steels was examined and compared with Nb-microalloyed and Ni-added Nb-microalloyed steels of comparable strength level intended for LPG pipelines and storage vessels, or for use in as oil pipelines in arctic environments to meet the demands of high toughness at low service temperatures.

Microalloyed steels with high nitrogen levels are regarded as having inferior toughness, due to the higher amount of free nitrogen in the HAZ matrix after welding. This is especially relevant in V-microalloyed steels where higher nitrogen contents are used to optimize the strength. Comparison has been made of the HAZ toughness in Nb-, Nb+Ni-, and V-microalloyed steels after various welding cycles. The Nb- and Nb+Ni-microalloyed steel exhibited excellent toughness with Charpy ITT\(_{273} \) between -100°C and -80°C. Almost equal ITTs were found in the V-microalloyed steel with
low nitrogen (50 ppm N). These low ITT$_{27J}$ values are due to a weld normalized HAZ with polygonal ferrite and pearlite. A high nitrogen V-microalloyed steel (130 ppm N) demonstrated increases in transition temperature compared to low nitrogen steel when welded with low (16 kJ/cm) and high (80 kJ/cm) heat inputs.

However, welding with intermediate energy (30 kJ/cm) resulted in the same ITT$_{27J}$ as for the low nitrogen steel. This was shown to depend on VN precipitation in the HAZ during welding of the subsequent pass. The level of free nitrogen in the HAZ matrix is then reduced at the same time as weld normalization occurs. Both these effects are beneficial to toughness. From the results it is shown that V-microalloyed steels with higher nitrogen levels can be successfully welded with excellent HAZ toughness, and thereby possible for service in pipelines, if proper welding conditions are selected.


In Paper F, the influence of nitrogen on the toughness of the weld heat affected zone has been examined for two HSLA steels. This is of interest in the context of SiCa-treatment, which is adopted in order to modify the inclusion morphology in steel and to lower the sulphur content in the steel. In general, this treatment leads to an increase in the nitrogen content. The main nitrogen uptake takes place during pouring of the steel from converter to ladle, while the increase in nitrogen during the actual SiCa-treatment is of the order of 0.001 - 0.003%. The nitrogen level after SiCa-treatment can be about 90-110 ppm, and in the worst cases up to 140 ppm. In the first part of this investigation, two SiCa-treated, niobium-micro-alloyed steels each with two different nitrogen levels were studied. By welding simulation and Charpy-V-testing, the influence of nitrogen on the brittle-to-ductile transition temperature was established; it was found that the transition temperature increased by 3-4°C/0.001%N for a steel alloyed with Nb+Ni and by 2-5°C/0.001%N for a straight Nb-micro-alloyed steel.

The second part of the investigation deals with LPG-grades containing 0.5% nickel (+Nb) and without nickel. The steels were welded using three different levels of arc energy; 12, 20 and 30kJ/cm whereupon Charpy-V-impact and COD-tests were performed for the HAZ. The aim of welding with three different arc energies was to vary the HAZ austenite grain size which increases with increasing arc energy. The results demonstrate that the nickel content did not have any great influence on the brittle-to-ductile transition temperature, determined via Charpy-V-testing, and that this temperature is for both deposits satisfactory. The COD-testing, however, revealed an increased level of toughness for the nickel-alloyed LPG-steel. It should also be pointed out that the δ$_c$-values for both steel-variants fulfilled a Japanese (12) specification for construction of LPG-tanks. The third and final part of the investigation presents results from a microstructural examination of the weld simulated niobium-micro-alloyed steel (+Ni), which was also included in the first part. The aim was to examine if differences in degree of precipitation could possibly be the reason for the differences in hardness between high and low nitrogen steels, since no major alterations of microstructure were observed in the weld simulated steels -with different N-levels. Electron-microscope
studies revealed that the amount of precipitated particles in the HAZ was considerably lower, about 1/20 of that in the base material. Moreover, the precipitates in the HAZ almost always contained titanium, which indicates that titanium nitrides have remained unsolved at the peak temperature, and functioned as nucleation sites for further precipitation of mainly Nb (C,N) during cooling. It has been revealed that re-precipitation of dissolved Nb (C,N) occurs only to a limited extent in the HAZ, and that these re-precipitated particles are too few to explain the differences in hardness. Therefore, it is of interest to study further the reasons for the observed loss of HAZ toughness and the increase in hardness, which are associated with an increased nitrogen level.


This work is one of a series in which the influence of manganese and silicon on the mechanical properties of weld metal deposits, especially toughness, are studied. Previous reports have discussed submerged arc welding and manual metal arc welding of mild steel as well as high-strength steel. In this study, the Mn/Si-ratio in the wires, and thereby the ratio of these elements in the deposit has been varied. Furthermore, two different shielding gases have been used, a mixture of 80 % argon and 20 % carbon dioxide (AK 20) and pure carbon dioxide (CO2). The alteration of shielding atmosphere was made in order to engender different levels of oxygen in the deposits. The results indicate an optimum Mn/Si-ratio, pertaining to maximum toughness, of 3 - 4 for AK 20 and ~ 4 - 7 for the C02-welded deposits. In this context, it should be noted that the optimum toughness appears at a manganese level of 1-1.2 %, independent of the shielding gas. Both lower and higher levels of manganese result in a deterioration in toughness. This optimum in manganese level can be explained by the lowering of the Ar-temperature derived from this element, which results in an enhanced nucleation of ferrite and thereby a grain refining effect. Increasing manganese also gives rise to a microstructural change, whereby the amount of acicular ferrite is increased at the expense of grain-boundary and widmanstätten ferrite. Moreover, increasing the amount of manganese, engenders a strong solution-hardening effect. The negative effect of the latter factor on toughness is counteracted at manganese levels below 1-1.2 % by the previously-mentioned microstructural modifications. Further increases in manganese-level above the optimum, however, result in a decrease in toughness.

The optimum silicon content will be determined by the optimum Mn/Si-ratio and the manganese level. It should be noted, however, that a silicon level sufficient to kill the weld deposit is required, i.e. > 0.15-0.20 %. The best toughness of the deposits examined (welded with AK 20) has been obtained for Mn = 1.09 % and Si = 0.34 % for which the brittle-ductile transition temperature is -90°C. For the CO2-deposits, excellent toughness was achieved, transition temperatures of -80°C and -81°C, at Mn = 0.97 %, Si = 0.25 % and Mn = 1.15 %, Si = 0.17 % respectively. The increased amount of oxygen in the CO2-welded deposits, as compared to the
AK 20-deposits, resulted in a greater average diameter of micro-inclusions, which in turn leads to a reduced level of upper-shelf energy.


The work presented in Paper H has been produced within the SEGA-co-operation and is concerned with the influence of the levels of manganese and silicon on the mechanical properties, especially toughness, of weld metal deposits. Hagström and Schölin carried out the metallographic evaluations, while Hansson made mechanical testing, scanning electron microscopy, evaluation and report writing.

The process elucidated in this report is plasma welding, while the preceding reports have discussed submerged-arc-welding of mild steel as well as high tensile steel, manual-metal-arc welding and MAG-welding with two different shielding gases (AK20 and CO2). The filler material used for the plasma welding contained different levels of manganese and silicon in order to vary the chemical composition of the weld deposits. Due to the fact that the plasma welds were made in 6 mm thick plates, compared with 25 mm plates in the former studies, SAW and MAG-welding of 6 mm plate were carried out in order to obtain comparative material. The plate material used had a base analysis 0.043%C, 0.19%Si and 0.23%Mn.

The present study has shown that the degree of interfusion by plasma-welding is about 85-90%, and thus the filler materials used are not capable of giving sufficient levels of manganese in the deposits to obtain an optimum microstructure from the viewpoint of impact toughness, i.e. a large proportion of acicular ferrite. Instead the microstructure consists mainly of coarse polygonal ferrite and the impact transition temperature is around 0°C for all deposits. PASEM-studies revealed that the area fraction of micro-inclusions in the plasma-welded deposits is considerably lower, about one tenth of that in the SAW-and MAG weldments.


Paper I describes the properties of wear resistant steels, the HARDOX grades produced by SSAB Oxelösund AB. Hansson has written the part with respect to the influence of alloy content on cutting and welding properties, as well as bending properties. These grades have good weldability, where especially the HARDOX 400 can be welded in combined thicknesses up to 50 mm without need for preheating. In thicker gauges HARDOX 400 needs a moderate preheat when welded. The HARDOX 500 which has a higher carbon content needs a preheat of 100°C when welded in combined thicknesses up to 35 mm. At higher combined thicknesses this grade needs a preheat of 150°C, which is quite moderate. Löwgren has written the part regarding wear of abrasive resistant steels.
Moulds for plastic fabrication are commonly produced using quenched and tempered tool steels with a nominal hardness of 300 HBW. After machining and surface preparation these moulds are ready to go into production. The chemical compositions commonly used in these steels have high carbon and alloy contents, which necessitates oil quenching from the austenitizing temperature. When producing moulds in higher hardnesses, above approx. 38 HRC, the traditional mould material is supplied in soft-annealed condition. After machining the tool is quenched and tempered to the hardness level desired. However, during the heat treatment operation shape changes often take place due to the:

- ferrite $\rightarrow$ austenite transformation during heating to the austenitizing temperature desired
- austenite $\rightarrow$ martensite transformation during the quenching, and
- the decomposition of martensite and precipitation reactions which take place when tempering the steel.

The changes are dependent on mould steel chemical composition, austenitizing temperature, cooling rate during the quench operation, mould geometry etc. Due to these dimensional changes the mould maker has to adjust the shape of the actual mould when it returns from the heat treatment. Furthermore, the cracking risk, which might be quite high due to the high carbon and alloy contents normally used in these steels, during the heat treatment operation should not be neglected. Today, the machining technique develops rapidly which gives possibilities to machine also hard materials. This has led to the development of a new generation of tool steels, recently introduced on the market, delivered as quenched and tempered by the steel producer. As these grades need no further heat treatment, a mould is ready to set into production when it is machined and, also surface treated if necessary. The shorter lead times possible in mould manufacturing make these steels interesting in technical as well as in economical terms. The new grades possess an excellent machinability, in combination with a far better weldability, as compared to the tool steels commonly produced for the same purpose.

High strength MAG solid welding wires with yield strength above 900 MPa are not to be found in the catalogues of the large consumables manufacturers. One reason is that nowadays small heats of steel cannot be manufactured due to cost consideration and also the need for such wires has not been urgent enough. Instead of a stronger wire it is possible to compensate with a larger leg length where necessary on fillet welds. On butt welding such a compensation is however difficult to realise. While developing a new tool steel therefore also an experimental wire was tested with yield strength of 1100 MPa with very low contents of both phosphorus and sulphur.
This new high strength steel wire was tested as a TIG welding wire for the new tool steel. Also the wire was tested with automatic TIG welding and MAG welding on structural high strength QT steel.

The paper refers to result from a rather broad investigation where the weldment toughness was studied as the primary parameter. At this high strength level the toughness will be critical and a low heat input is important, as well as shielding gas to obtain a very clean weld metal as to oxygen and sulphur inclusions.

In many structural applications the critical flaw size is the parameter of the largest interest with these high strength levels. In tool steels the expectations of critical flaws are in general weak. This is explicitly attributed to the inherent low fracture toughness of this category of steel as well as to the high design stress that can be used. The impact of differences in tool steel toughness on critical flaws is shown using the well established BS method for fitness-for-purpose evaluation. From the obtained Charpy toughness values the $K_{\text{mat}}$ fracture toughness values were calculated. Improvement in HRC 33 tool steel base metal is shown as well as the limitations when assessing flaws in tool steels weldments.


This is the Swedish Patent regarding chemical composition and processing Requirements in the production of the two new steels developed.
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