

# Weld Metal Properties

for

## Extra High Strength Steels

by

Kenneth  
Håkansson



**Doctoral Thesis**

**Report 2002 - August**

**Division of Welding**

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## Preface

The research presented in this thesis started 1989 and has mostly been carried out at the Laboratory at Kockums AB in Malmö and some of the work at Aeronautical and Maritime Research Laboratory (AMRL) in Melbourne, Australia which belongs to Australian Department of Defence, Defence Science and Technology Organisation.(DSTO) During 1989 to 1990 I had an exchange scientist from AMRL at Kockums and we worked together doing welding research of common interest regarding welding of submarines. From April 1991 and to the end of March 1992 I was an exchange scientist at AMRL continuing the work we started at Kockums. I am still working with research regarding welding of extra high strength steels for submarines.

The work has been financed by Kockums AB and they funded the research performed at AMRL.

I wish to express my sincere gratitude to Prof. Nils Erik Hannerz at division of Welding, Department of Industrial Production at The Royal Institute of Technology (KTH).

To Dr B Göran Johansson at Kockums AB who encouraged me to write this thesis and to Kockums , and their personnel especially at the Laboratory. I owe many thanks to Kjell Nilsson welding the submerged arc test plates, to Bo Palm welding the manual metal arc test plates, Gert Andersson, Tommy Svensson, Bo Lökvist machining the test specimens and to John Ivarsson testing the specimens.

I would like to thank Lars-Eric Larsson, Arne Torstensson and Berth Eriksson at Kockums AB, for discussions around fatigue and design.

Thanks also to Kjeld Borggren at Sydtek AB for valuable comments and fruitful discussions around the work.

I would like to thank DSTO and their personnel especially Dr Brian Dixon who is co-writer in some cases and Dr John Ritter both at AMRL for fruitful collaboration and discussions.

Thanks also to Dr Bo Janzon at Försvarets Forsknings Anstalt (FOA) now FOI, for discussions around deformation and strain rate during explosive loading.

Staffanstorp, August 2002

Kenneth Håkansson



## ABSTRACT

Weld metal properties of extra high strength steels has been investigated in order to find the optimum welding parameters to achieve properties fulfilling the requirements put on structures in large thickness. The increased demand for higher performance of welded structures has put the focus on weight reduction and one way of doing this is to use steels with higher yield strength. Matching mechanical properties in the weld metal has so far been difficult to obtain as the 0.2% proof stress has increased for the parent steel. This put limitations on the heat input, on the preheat as well as the interpass temperature. Too high heat input as well as high preheat and interpass temperature will result in low strength, low hardness, and low toughness, while low heat input or low preheat and interpass temperature will give risk for hydrogen cracking. The limitations of the welding parameters are greater as the strength of the parent metal increases

The optimum combination of welding parameters, notably preheat and interpass temperature, was identified for manual metal arc welding of a 690 MPa quenched and tempered steel to give maximum HAZ and weld metal toughness. Results showed that the minimum preheat and interpass temperature for control of weld metal hydrogen cracking was in this case 60°C. Low values of toughness in both weld and HAZ consistently occurred in the weld root region. Welding technique involving low preheat temperature and one-sided welding should therefore be avoided.

Improving the productivity in fabrication in 690 MPa quenched and tempered steel constructions welded by submerged arc process was done with the addition of alloyed metal powder. Despite using the standard techniques for control of hydrogen cracking, transverse cracks were identified.

Submerged arc welding of a 900 MPa strength steel shows that the weld metal is not achieving the same 0.2% proof stress as the parent metal in spite of very highly alloyed filler material. Hydrogen cracking in the form of transverse cracks in the weld metal was observed when the preheat temperature was below 200°C. Some of the welds were welded with cold wire addition.

Several test methods for simulation of hydrogen cracking exist, but only a few of them are suitable for submerged arc welding. Most of the techniques for testing of hydrogen cracking involve only one single pass. Techniques for testing for hydrogen cracking in multipass submerged arc welding have for this reason been developed.

The weaving is often restricted in manual metal arc welding procedurs in order to avoid excessive heat input. The weaving pattern has a pronounced influence on the toughness in extra high strength steels.

Post weld heat treatment of welded structures is performed in order to reduce the residual stresses created by the welding. This thermal stress relief has a detrimental effect on the toughness of the weld metal in both manual metal arc welds and submerged arc welds in extra high strength steels.

Safe welding procedures should result in welds with mechanical properties meeting the same requirements for strength and toughness as for the parent material and at the same time be free from defects, in particular cracks. The defects in the form of transverse cracks found in the weld metal are of a size about 5 mm x 10 mm, and they are often embedded in the weld. This kind of flaws is not easily found by the non-destructive testing. Techniques for assessing the flaws has been used in order to get a requirement for toughness for welded joints in extra high strength steels subjected to impact loading and fatigue loading.

## KEY WORDS

Welding parameters

Toughness

Hardness

690 MPa Steel

900 MPa Steel

Preheat temperature

Interpass temperature

Heat input

Strain rate

Fracture toughness

Defect size

Hydrogen cracking test

Mechanical properties

Thermal stress relieve

Metal powder addition

Cold wire



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Paper A: *Effects of Welding Parameters on Weld Zone Toughness and Hardness in 690 MPa Steel.*

Paper B: *Submerged Arc Welding with Alloy Powder Additions for high Strength Steels.*

Paper C: *The Influence of Wide weld Beads on the Mechanical Properties for Manual Metal Arc Welded High Performance Steel.*

Paper D: *Effect of Post Weld Stress Relief Heat Treatment on Charpy Impact Toughness of 690 MPa Yield Stress Weld Deposits.*

Paper E: *Personal Reflections on 20 Years of R&D in shipbuilding – Its Implementation and Future Aspects.*

Paper F: *Development of Hydrogen Cracks in High Strength Submerged Arc Weld Metal- An Investigation by Factoral Analysis.*

Paper G: *Estimation of Fracture Toughness According to Sanz's and Wallin's Model, in Electrodeposited Welds of about 25 Year Old Ships.*

Paper H: *Estimation of Strain Rate at Explosive Loading by Wallin's model and Data from Explosion Bulge Test, Flaw bulge Test and Charpy-V Notch Test to Calculate Flaw Size and Impact Requirements.*

Paper I: *Submerged Arc Welding of WELDOX 900E in 50 mm Plate Thickness-Hydrogen Cracking Tests.*

Paper J: *Submerged Arc Welding of WELDOX 900E in 50 mm Plate Thickness- Cold Wire Addition and Mechanical Properties.*

The author's views are limited to the author.

# DISSERTATION

## **Paper A**

B. Dixon and K. Håkansson., *Effects of Welding Parameters on Weld Zone Toughness and Hardness in 690 MPa Steel*, Paper published in *Welding Journal*, Volume 74, No 4, April 1995, page122-s to 132-s.

## **Paper B**

K Håkansson and B. Dixon., *Submerged Arc Welding with Alloy Powder Additions for high Strength Steels*, Paper published and presented in 42<sup>nd</sup> WTIA National Welding Conference 24 to 27 October 1994, paper 21. Paper also published and presented at The International Conference on Joining of Materials JOM-7, may 31 to June 2, 1995 in Helsingør, Denmark, page 280 to293. Paper also published in *International Journal Joining Material* 1996, vol. 8, no. 1, pp 14-21.

## **Paper C**

K. Håkansson, *The Influence of Wide Weld Beads on the Mechanical Properties for Manual Metal Arc Welded High Performance Steels*, Article made from Kockums AB Internal Report 16/60/1239E, page 1 to 10, April 1995. 12 pages.

## **Paper D**

K. Håkansson and B. Dixon., *Effect of Post Weld Stress Relief Heat Treatment on Charpy Impact Toughness of 690 MPa Yield Stress Weld Deposits.*, Paper published and presented at The ASM International European Conference on Welding and Joining Science and Technology, Madrid, Spain, 10 to 12 March 1997, page 593 to 607.

## **Paper E**

K. Håkansson, *Personal Reflections on 20 Years of R&D in Shipbuilding – Its Implementation and Future Aspects*, Paper published and presented at The International Conference on The Joining of Materials JOM-8, 12 to 14 may 1997 in Helsingør, Denmark, page 11 to 19.

## **Paper F**

K. Håkansson and M. Dahlander, *Development of Hydrogen Cracks in High Strength Submerged Arc Weld Metal- An Investigation by Factoral Analysis*, Paper published and presented at The International Workshop on Hydrogen Management for Welding Applications, 6 to 8 October 1998, page 259 to 282.

## **Paper G**

K. Håkansson, *Estimation of Fracture Toughness According to Sanz 's and Wallin 's Model, in Electro gas Welds of about 25 Year Old Ships*. Paper submitted to IIW delegate for publication.

**Paper H**

K. Håkansson, *Estimation of Strain Rate at Explosive Loading by Wallin's model and Data from Explosion Bulge Test, Flaw bulge Test and Charpy-V notch Test to Calculate Flaw Size and Impact Requirements*. Invited lecture given the 21 of March 2002 at Welding Technological Institute of Australia, Panel no 2: Welding Metallurgy.

**Paper I**

K Håkansson, *Submerged Arc Welding of WELDOX 900EM in 50 mm Plate Thickness-Hydrogen Cracking Tests*.. Paper published and presented at the International Conference on " Technological and Research Developments in Welded Defence Equipment" 18-19 March 2002, Melbourne, Australia.

**Paper J**

K. Håkansson, *Submerged Arc Welding of WELDOX 900EM in 50 mm plate Thickness-Cold Wire Addition and Mechanical Properties*. Invited lecture given the 21 of March 2002 at Welding Technological Institute of Australia, Panel no 2: Welding Metallurgy.

## **DIVISION OF WORK BETWEEN AUTHORS**

### **Paper A**

This co-operative work was performed when Dr Brian Dixon was working as exchange scientist at Kockums AB. K. Håkansson initiated the work in the part regarding welding procedures and properties. The experimental work was planned and directed by the author and Dr B Dixon together.

The part involving hardness measurements on bead on plate welds was planned and directed by Dr B Dixon. K. Håkansson took part in the discussion of results and the main conclusions. Dr B Dixon prepared the paper after he returned to Australia.

### **Paper B**

This work was done when K. Håkansson was working as an exchange scientist at Defence Science and Technology Organisation, Aeronautical and Maritime Research Laboratory, Melbourne, Victoria, Australia. K. Håkansson designed and performed the experiments. The results were evaluated and discussed together with Dr Brian Dixon.

K. Håkansson wrote the report.

### **Paper D**

This work was done when K. Håkansson was working as an exchange scientist at Defence Science and Technology Organisation, Aeronautical and Maritime Research Laboratory, Melbourne, Victoria, Australia. K. Håkansson designed and performed the experiments. The results were evaluated and discussed together with Dr Brian Dixon

K. Håkansson wrote the report.

### **Paper F**

K. Håkansson initiated and planned the investigation. Mats Dahlander performed the experimental work, under supervision of K. Håkansson. Mats Dahlander wrote a report in Swedish in which K. Håkansson took part in the discussions and conclusions.

K. Håkansson wrote the paper in English and presented it at the workshop in Canada.

## NOTATIONS AND SYMBOLS

$CE_{IIW}$ Welding	carbon equivalent formula according to International Institute of
CET	carbon equivalent formula according to SS-EN 1011-2
E	welding arc voltage [V]
E	Young's modulus of elasticity [MPa]
I	welding current [A]
Q&T	Quench and Tempered
S	welding speed [mm/min]
$P_{cm}$	carbon equivalent formula according to Ito-Bessyo
$P_{HA}$	cracking parameter
HAZ	Heat Affected Zone
$T_p$	preheat temperature [°C]
$T_{PCET}$	preheat temperature due to CET [°C]
$T_{Pd}$	preheat temperature due to thickness [°C]
$T_{PHD}$	preheat temperature due to hydrogen content [°C]
$T_{PQ}$	preheat temperature due to heat input [°C]
d	thickness [mm]
HD	ml hydrogen per 100 g deposit material
Q	heat input [kJ/mm]
BIS812EMA	Q&T Steel made by Bunge in Australia
WELDOX 700EM	Q&T steel made by SSAB in Sweden
HY100	Q&T steel according to MIL-STD

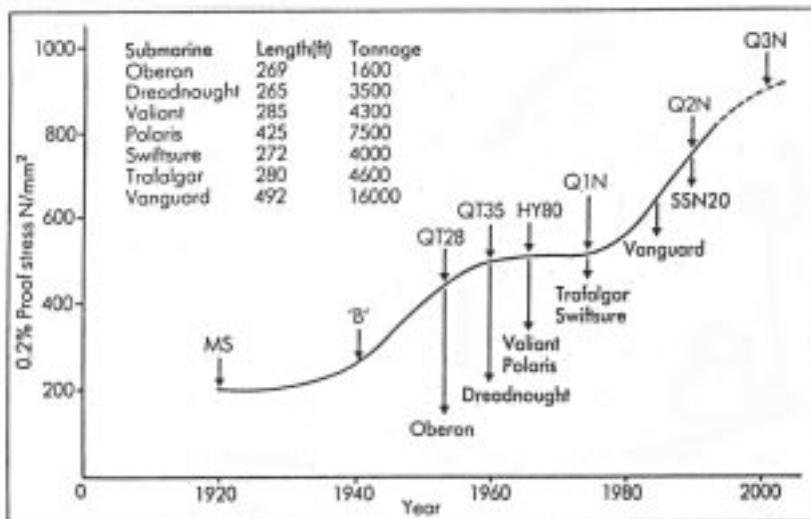
120S	solid wire used for submerged arc welding
WELDOX 900	Q&T steel made by SSAB in Sweden
SPOOLARC 120	solid wire used for submerged arc welding
SPOOLARC 140	solid wire used for submerged arc welding
OK Tubrod 15.29S	tubular wire used for submerged arc welding
CTOD	Crack Tip Opening Displacement
ROL	run out length
TWI	The Welding Institute in United Kingdom
$\dot{\epsilon}$	d $\epsilon$ /dt strain rate
D	diameter [mm]
ws	wire feed speed [m/min]
A	weld bead area [mm <sup>2</sup> ]
B	number of electrodes per kg weld metal
v	travel speed [mm/min]
$\delta$	density [kg/dm <sup>3</sup> ]

# 1.0 INTRODUCTION

## 1.1 Background

Quenched and Tempered (Q & T) low alloy steels are being used increasingly in both structural and abrasion resistant applications. The continuing increase in usage is the appreciation of the advantages these steels can offer when selected and fabricated and used correctly. Advantages arising from the use of Q & T steels relate to lighter structures, greater load carrying capacities and improved service life. High strength structural Q & T steels are mainly used in bridges, buildings, storage tanks, equipment as mobile cranes, road and rail tankers and submarines etc, where loading are high and weight saving is critical. All these products involve the demand for high performance steels and joining methods together with fabrication methods which preserve the properties of the steel. The increased performance has put the focus on weight reduction. One way to get weight reduction is to use steels with higher yield strength. This has led to a need for improved knowledge of the influence of the manufacturing process on these high strength steels. A very important manufacturing process for these extra high strength steels is the welding process. The history of the development of high strength steels for submarine and surface vessel construction are shown in figure 1. [1]

Figure 1. The development of 0.2%Proof stress for steels used in submarine design since 1920.



It needs to be pointed out that as the yield point levels are increased, the ultimate tensile strength cannot be increased in proportion. Therefore the yield to ultimate ratio of steels for naval use is high compared with that for conventional structural steels, This means greater demands on the accuracy of the stress design as the corresponding factor of safety is reduced. The development of steels has been so successful that the properties of the welded joints have been unable to keep pace. [1, 2, 3, 4].

In order to achieve the optimum results in the weld metal and heat affected zones of welds in these Q & T steels, stringent requirements exist for level of preheat, the hydrogen content of the consumables, the heat input controls, and the specific bead placement techniques.



The following areas are important for the development of welding procedures in general and for extra high strength steel in particular. The relatively high hardenability necessary for development of required mechanical properties by hardening and tempering renders Q & T steels more susceptible to the thermal effects of welding than conventional carbon and carbon-manganese structural steels. Depending on weld thermal cycle, the metallurgical changes due to welding may range from overtempering, resulting in loss of strength, hardness and possibly toughness, to the formation of undesirable microstructures causing hardening and embrittlement.

Use of correct welding procedures is therefore of fundamental importance, and when proper procedures are employed Q & T steels can be welded satisfactorily by all major welding processes.

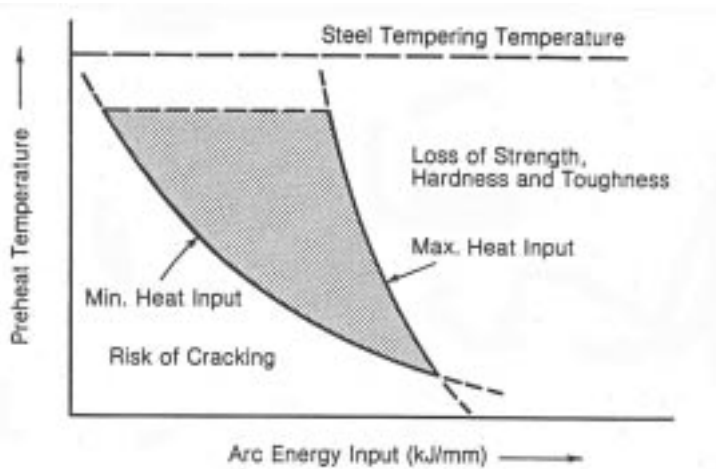
Because the Q & T steels differ from conventional steels, welding procedures and practices are more stringent due to limitations in tempering temperature, which can create softening in the HAZ. In particular it is necessary to observe limitations on both maximum and minimum total weld heat input, as illustrated schematically in Figure 2. [5]

Note that the concept of total weld heat input involves consideration of:

- Preheat temperature
- Interpass temperature
- Arc energy input  $Q = E \times I / S \times 60 / 1000$

In the case of alternating current RMS is used for the arc voltage and welding current.

Figure 2. Total Weld Heat Input Relationships for Welding Q & T Steels



The shaded area is defined by the curves in figure 2 denotes a permissible heat input "window" limited on the low energy side by the risk of excessive hardening and cracking. On the high energy side the heat input window is limited by loss of strength and hardness due to overtempering and possible loss of toughness due to re-transformation on cooling to upper bainitic microstructures.

The wedge shape of the window favours the use of low to moderate arc energies together with substantial levels of preheat in order to secure adequate tolerance to total heat input variations likely to occur under practical shop or field welding conditions.

## 1.2 Carbon equivalent formulas

When welding high strength steels preheat of weld preparations is necessary for the control of hydrogen (cold) cracking, to avoid excessive hardness in weld heat affected zones, and to provide adequate toughness for shock resistance. The required levels of preheat may be determined by any of a number of techniques, most of which are based upon empirical relationships between base metal composition, as a carbon equivalent, weld restraint, microstructure and cooling rate.

BS 5135:1984 [6] for example, provides a minimum preheat requirement to control cracking as a function of the composition of the steel, the plate thickness in form of combined plate thickness, the hydrogen level for the filler material and the heat input.

One problem with these established relationships is that they were generally developed between 1950 and 1980, and therefore, relate to more popular steels available during that period. For example, the influence of composition is commonly measured by calculating a carbon equivalent (CE) value according to a formula such as the International Institute of Welding (IIW) formula [7]:

$$CE_{(IIW)} = \%C + \%Mn/6 + (\%Cr + \%V + \%Mo)/5 + (\%Ni + \%Cu)/15$$

This simple formula does not take into account the influence of small quantities of the highly potent alloying elements such as Nb, B and N, which may be used in modern steels. The formula describes the hardenability of the steel. Several other carbon equivalent formulae have been proposed from time to time. Normally these carbon equivalent formulas were used for describing hydrogen cracking in the HAZ. Many of them have not been extensively used, because of their unfamiliarity, their excessive complexity, or because they were almost identical to the IIW formula. Another carbon equivalent formula developed in Japan for steels with low carbon content is the Pcm formula:

$$P_{cm} (\%) = C + Si/30 + Mn/20 + Cu/20 + Ni/60 + Cr/20 + Mo/15 + V/10 + 5B. [7]$$

Furthermore, the sensitivity of a steel to cracking may be influenced by factors other than composition, thickness and heat input. For example, the base material strength and the filler material used influence the level of restraint across the weldment, which may be dependent upon the prior thermal and mechanical treatments in addition to the chemical composition. Other factors that may have a significant influence on hardness and sensitivity to hydrogen cracking include the joint design, weld metal hydrogen content and the amount of diffusible hydrogen in the base material.

A development from the Pcm formula is the  $P_{HA}$  cracking parameter.[8].  $P_{HA}$  is based on the analysis of cases of hydrogen induced cracking in welded high strength steels by Tekken type test, and establishes a critical preheating temperature necessary to avoid cracking in HAZ and in the root of the weld. As a general rule, for both carbon-manganese and low alloy steels, the harder the microstructure the greater risk for cracking.

Many test methods have been developed for evaluating the risk for cold cracking correlated to the restraint and hardness in different zones in a welded joint. [7-14]

A carbon equivalent formula that takes special care of cracking in the weld metal is the CET formula. If the carbon equivalent CET of the parent metal exceeds that of the weld metal by at least 0.03%, the chemical composition for the parent steel is used in the calculation.[15] Otherwise the calculation of the preheat temperature has to be based on the CET of the weld metal increased by 0.03%

The carbon equivalent formulas were strictly developed for determining the hardenability of HAZs. Their adaptation to hydrogen cracking in welds and HAZs has developed out of engineering practice.[16] Furthermore, these formulas can not be used for other materials than these for which they were developed.

### 1.3 Calculation of the preheat temperature for different steels and consumables

The calculation is performed, by using the new standard as a reference SS-EN 1011-2 "Welding – Recommendations for welding of metallic materials – Part 2: Arc welding of ferritic steels".[15].

Annex C Avoidance of hydrogen cracking (also known as cold cracking).

This standard and this part is the so far the only method for calculation of the preheat temperature for weld metals.

C.3 Method B for the avoidance of hydrogen cracking in non-alloyed, fine grained and low alloy steels.

Carbon equivalent (CET) =  $C + (Mn+Mo)/10 + (Cr+Cu)/20 + Ni/40$

It applies to the following range of concentrations (per centage by weight)

Carbon	C	= 0.05 to 0.32
Silicon	Si	= max. 0.8
Manganese	Mn	= 0.5 to 1.9
Chromium	Cr	= max. 1.5
Copper	Cu	= max. 0.7
Molybdenum	Mo	= max. 0.75
Niobium	Nb	= max. 0.06
Nickel	Ni	= max. 2.5
Titanium	Ti	= max 0.12
Vanadium	V	= max. 0.18
Boron	B	= max. 0.005

#### 1.3.1 Calculation of the preheat temperature

If the effects of chemical composition, characterised by the carbon equivalent CET, the plate thickness d, the hydrogen content of the weld metal HD and the heat input Q, are combined the formula given below can be derived which can be used to calculate the preheat temperature  $T_p$ .

$$T_p = T_{pCET} + T_{pd} + T_{pHD} + T_{pQ}$$

$$= 697 \times CET + 160 \times \tanh(d/35) + 62 \times HD^{0.35} + (53 \times CET - 32) \times Q - 328$$

This relationship is valid for

CET	= 0.2 % to 0.5 %
d	= 10 mm to 90 mm
HD	= 1 ml/100g to 20 ml/100g
Q	= 0.5 kJ/mm to 4.0 kJ/mm

for structural steels with a yield strength up to 1000 N/mm<sup>2</sup>.

According to experience, the preheat temperatures calculated are valid provided that the following conditions are fulfilled.

- The carbon equivalent CET of the parent metal exceeds that of the weld metal by at least 0.03%. Otherwise the calculation of the preheat temperature has to be based on the CET of the weld metal increased by 0.03%
- Single-pass fillet, tack and root welds have a minimum length of 50 mm. If the plate thickness exceeds 25 mm, tack and root passes are deposited in two layers using a mild ductile weld metal.
- In the case of filler pass welding, which also includes multipass fillet welds, no interpass cooling takes place as long as the weld thickness has not yet attained one third of the plate thickness. Otherwise, it is necessary to reduce the hydrogen content by means of a post weld heat treatment.
- The welding sequence shall be selected in such a way that the strong plastic deformations of the only partly filled welds are avoided.

The chemical composition is given as typical values found during different work.

	BIS 812 EMA	WELDOX 700EM	HY100	Cast Steel	SAW con- sumables 120S	MMA Atom Arc 12018M2	WELDOX 900
C%	0.13	0.12	0.15	0.09	0.09	0.04	0.16
Mn%	0.93	0.91	0.42	0.37	1.54	1.55	0.85
Mo%	0.39	0.45	0.50	0.50	0.47	0.26	0.60
Cu%	0.21	0.19	0.01	0.00	0.01	0.00	0.20
Ni%	1.28	1.35	3.02	4.40	2.36	3.67	1.86
CET%	0.328	0.325	0.393	0.327	0.335	0.313	0.391

For a simple comparison it is possible to choose the thickness to be 29 mm, the hydrogen level to be 3.6 ml / 100 g and the heat input to be 1.0 kJ/mm. For such a case the calculation will be as follows:  $T_{pCET} = 750 \times CET - 150$ .

Which will give following preheat temperatures:

Preheat Temperature	BIS 812 EMA	WELDOX 700EM	HY100	Cast steel	SAW consum- ables 120S	MMA Atom Arc 12018M2	WELDOX 900
°C	96	94	145	95	101*	85**	143

The preheat temperature when the composition for the weld metal is considered is shown below.

\*Due to a higher carbon equivalent for the filler material compared to the parent steel (WELDOX 700EM) an addition of 0.03 to CET has to be done because the filler material (120S) has higher CET than WELDOX 700EM. The preheat will be 124 °C instead for 101°C.

\*\* WELDOX 700EM do not exceed the weld metal with CET 0.03. 0.343 is used in the calculation instead of 0.313. The preheat is calculated to be 107 °C instead for 85°C.

#### **1.4 Inter Pass Temperature and cooling rate**

Hydrogen cracking and embrittlement of ferritic steels occurs only at ambient temperatures. It is therefore possible to avoid cracking in a susceptible microstructure by maintaining it at a temperature high enough to allow the hydrogen to diffuse away. Any measure, which slows down the cooling of the weld, is therefore helpful in reducing the hydrogen level. In fabrication the cooling rate is normally controlled by varying the heat input and by the preheat temperature.[7]

The microstructure can be made softer by tempering, which can be done by using multipass welding, temper bead technique and post weld heat treatment. An increase in temperature will also increase the rate of diffusion of hydrogen.

Normally the hardenability of the steel and in most case also the HAZ, is governed by its composition, which can be established by a carbon equivalent formula.

For weld metal microstructure the situation is more complex, because the toughness of weld metal does not depend on hardness in the same way as HAZ toughness does. A tough weld metal microstructure appears to be better able to resist hydrogen cracking than one, which is less tough. In a relatively hard, alloyed weld metal the cracking is usually transverse and perpendicular to the weld surface.

#### **1.5 Avoiding flaws in the weld metal**

Hydrogen-induced cracking occurs when the following conditions occur simultaneously. [7, 16]

- Hydrogen is present to a sufficient degree.
- Tensile stresses act on the weld
- A susceptible microstructure is present

With the introduction of modern-alloyed steels, the problem of HAZ cold cracking has been virtually eliminated. As a consequence the weld deposits have become critical and the regions of the joint most prone to cracking and embrittlement.[4]

Welding of steels with yield strength from about 600 N/mm<sup>2</sup> and higher may often involve a filler material which is higher alloyed than the steel itself in order to give a matching yield strength of the weld metal. Often the preheat or the inter pass temperature is calculated or derived from the chemical composition of the steel to avoid hydrogen cracking in the heat affected zone (HAZ). However, the weld metal can have a higher risk for cracking and the cracking may occur as transverse cracks or chevron cracks. The transverse cracks in the weld metal do not always reach the surface of the weld. They can be buried inside the weld which make them difficult to detect by non destructive testing.

Furthermore, control of cracking may not be the only requirement of a given welding procedure. In the fabrication of naval vessels, it is essential that the hull has adequate toughness to withstand shock loading for instance caused by exploding mines or torpedoes. It is important to keep the hydrogen level low in the weld metal. Electrodes and fluxes used for welding high strength steels needs to have low moisture absorption during storage or must be stored to avoid moisture. Electrodes and fluxes should be capable of being redried, if necessary, prior to use. Fluxes for submerged arc welding should be stored in heated hoppers and transferred to a heated hopper on the welding machine. If recirculation of flux by compressed air is used, the air must be free from moisture and oil. This can be done by having a filter to clean the compressed air from such contaminants.

Rust on the filler material or on the surfaces in the joint or close to the joint will generate hydrogen, which can be absorbed in the weld metal. Hydrogen absorption during welding can lead to cracking in the heat affected zone or in the weld metal.

## 1.6 Hydrogen level

Hydrogen is introduced into the weld during the welding process, from the atmosphere, from hydrocarbons on the material being welded, or from using electrodes or flux containing moisture, which depends on what type of flux that is used. [16-18]

Compared with other interstitials (carbon and nitrogen), hydrogen diffuses much more easily through b.c.c iron. An interesting feature of hydrogen diffusivity in ferritic steels is the wide scatter observed at temperatures below and around 200 °C.[7] The reasons for the simultaneous diffusivity reduction and increased scatter below 200 °C, is thought to be that hydrogen becomes trapped at various discontinuities where it transforms to molecular hydrogen. Hydrogen diffuses out from the weld, with a velocity depending on the temperature of the weld. [7, 10].

The bead size is also affecting the amount of hydrogen present in the weld metal, a large weld will have a higher hydrogen level compared with a small weld due to a larger diffusion distance. [7, 10]. It is possible to build up higher and higher hydrogen levels in multipass submerged arc welds, which normally has a large bead size. If the productivity is increased by metal powder addition, there is one another source for hydrogen, and at the same time the bead size is increased, which means that the risk for hydrogen cracking is greater.

## 1.7 Stress level

The thermal cycle generates contraction of the weld during cooling and builds up strains in the joint both due to the thermal expansion and due to transformation from austenite to ferrite or martensite. At a temperature of about 700 °C the stress is practically zero and considerable plastic strain may occur. The decrease in temperature is accommodated in the section, as elastic tensile stresses when ambient temperature has been reached. Depending on the rigidity of the structure, the residual stress can reach the yield strength of the metal. The combinations of these residual stresses and the residual strains imposed at high temperature may result in distortion. In rigid structures the natural contraction stresses are intensified because of the restraint imposed on the weld by the different parts of the joint. These stresses can be concentrated at notches such as the weld toe or the root of a weld.[7, 16, 19] In the weld metal, inclusions and other defects can act as stress and strain concentrators. High degree of strain gives high risk for cracking for a susceptible microstructure. The stress acting on a weld is a function of weld size, joint geometry, fit up, external restraint and the yield strength of the parent steel and weld metal.

## 1.8 Microstructure and properties of weld metals

The solidification or phase transformations during the thermal cycle of the weld are most likely to influence the final microstructure of the weld metal and the microstructure in HAZ after cooling to ambient temperatures. Some factors, which may have influence on the microstructure and properties, include: [7, 16, 19]

- The welding process itself, which determines the weld, pool size and geometry.
- The final composition of the melt as influenced by the filler material, the base material, fluxes, gas, moisture and how clean the joint was.
- The speed of welding and its effect on solidification speeds, crystal growth and segregation.
- The weld thermal cycle and its influence on microstructures produced during cooling.
- The effect of weld metal composition, on precipitation reactions especially in multipass welds

The cooling rate depends on the preheat, the thickness and the geometry of the parts to be joined and the heat input..

Controls over cooling rate in a particular fabrication is therefore achieved by varying heat input and preheat temperatures. The cooling rate has an influence on the mechanical properties of the weld metal and HAZ.

The cooling rate that is necessary to reach for the weld depends on the composition of the filler material used and the amount of hydrogen in the weld metal.

Welding of structures with small thickness, the cooling rate is normally slow, which decrease the mechanical properties i.e. the 0.2% proof strength and the impact properties. A recommendation for welding of the high strength steels is to use under matching filler material. In this case the risk for hydrogen cracking is decreased due to reduced stress level in the weld and HAZ. [20-22]

Also stainless filler material is recommended in same very difficult situations, where the austenitic weld metal can absorb the hydrogen and in this way reduce the risk for cracking. The mechanical properties for austenitic weld metal are around 500 MPa in 0.2% proof stress. The impact properties at low test temperatures is normally excellent. No preheat is necessary when welding with austenitic consumables.

The chemical composition governs the hardenability, and the traditional way of calculating the hardenability is to use empirical formulas, which defines a carbon equivalent. Most carbon equivalent formulas have been developed for the parent metal and no one so far has been used for the weld metal.

### **1.9 Hydrogen removal**

The time for hydrogen to diffuse can be calculated by Fick's second law and is described mathematically by the general differential equation: [7]

$$\delta C/\delta t = D \nabla^2 C.$$

This law describes the relationship between hydrogen concentration in the steel, C, time t, diffusivity D and spatial distribution characterised by  $\nabla^2$ .

From this it is possible to calculate and make hydrogen removal curves for different thickness and temperatures. The thickness of a fillet weld can be simulated by a cylindrical geometry. T butt, cruciform butt and corner butt welds can be regarded in this case as a fillet weld, and a butt weld can be simulated by plane plate geometry. [7]

### **1.10 Diffusion of hydrogen**

The diffusion of the hydrogen from the weld into the steel and out to the atmosphere depends on the temperature and the composition of the steel.

When welding with more than one bead the diffusion distance will increase and so will the amount of hydrogen for each consecutive bead. For multipass welded joints it is possible to get high amounts of hydrogen in the weld despite very low hydrogen welding consumables. This is due to the build up of hydrogen for each consecutive welding pass and at the same time the increase in diffusion distance. The diffusion of hydrogen is lower than the amount that is fed into the weld by each welding pass.

### **1.11 Weld bead size**

Heat input and productivity, here the deposition rate are related in such a way that high productivity means high heat input and at the same time a large bead. Larger bead size means longer diffusion distance for the hydrogen and also that more hydrogen is transferred into the weld.[23-25] On the other hand, the welding generates more heat, which increases

the diffusion of hydrogen. The bead size (A) cross sectional area in mm<sup>2</sup> for welding methods using solid wire with the diameter (D) in mm can be calculated according to following expression: in here the wire feed speed (ws) is expressed in m/min and the travel speed (v) is expressed in m/min.

$$A = (ws * \pi * D^2) / (4 * v)$$

The wire feed speed depends on the welding current and higher wire feed speed increases the nugget area.

For manual metal arc welding the bead size (A) in mm<sup>2</sup> can be calculated by using figures as number of electrodes per kg weld metal (B), the run out length (ROL) in mm and the density ( $\delta$ ) in kg/m<sup>3</sup> of the weld metal.

$$A = 10^9 / (B * ROL * \delta)$$

The only parameter which the welder can have an influence on is the ROL, and the ROL is directly affecting the bead size and the heat input. If the ROL is increased the heat input is reduced and so is the bead size.

### **1.12 Calculation of critical defect size**

The flaws found in high strength weld metals are hydrogen cracks are transverse type. Due to difficulties to find these defects and to measure their size one philosophy is to accept that there are always these kind of flaws of a certain size present in the weld. The size of the acceptable flaw depends on the impact properties of the material, in this case the weld metal, the estimated strain rate that the structure will experience during its life, and under which reliability the non-destructive testing is performed. Up to 50% of the flaws present can be missed during an ordinary non-destructive investigation.

The orientation of the flaw is important for both radiographic inspection and ultrasonic inspection. In radiographic inspection, the film will not reveal a thin flaw. If the flaw is orientated at an angle to the beam it will appear as a thin shadow not visible as a density difference in the film. For ultrasonic testing it is important to use the right type and orientation of the probe such that the sound can be picked up and not reflected away. Further problems should be taken into account such as ultrasonic testing for flaws close to the surface of thicker plates.

#### **1.12.1 Critical defect size due to fatigue**

The critical size and acceptance levels for defects in fusion welded joints can be calculated according to BS 7910 (former PD 6493). [26-28] The calculation programs Crackwise and Fatiguewise calculate example of flaws, found during experiments. (The latest version from TWI in UK). These programs are based on the recommendations given in PD6493. (Now BS7910)

All flaws with the size and type found in the welds were calculated to be acceptable under the assumed conditions.

The welding material can improve the fatigue strength. If the transition from austenite to martensite starts at lower temperatures, about 180°C, than the normal temperature of 400°C, the weld metal expand and induces compressive residual stress, which will increase the fatigue limit.[29].

#### **1.12.2 Calculation of critical defect size according to impact loads**

Structures may during its life experience different kinds of loads and stresses. Ships can be sailing during stormy weather and subjected to impact loading with a strain rate  $\epsilon$  of about



$10 \text{ s}^{-1}$ , while submarines can be subjected to depth charges which give a strain rate  $\epsilon$  of about  $10^4$  to  $10^6 \text{ s}^{-1}$ . [30-32]

These strain rates have an effect on the behaviour of the steel and the weld metal. Higher strain rate makes the material more prone to fracture in a brittle manner. Larger flaw size increases the risk for brittle failure. To be able to accept a certain flaw size in order to avoid brittle fracture due to impact loading a certain requirement has to be fulfilled.

In the present work a flaw size of 5 by 10 mm of a half elliptical surface breaking flaw is regarded as the size which is acceptable. This given size of flaw is without any safety margins. Flaws with larger dimension are supposed to and must be found by the non-destructive testing. The size of an imbedded flaw can be slightly larger

Depth charges and other explosives have been reported to give a strain rate in the same region as the one calculated from the explosion bulge test, about  $10^6$ .

It is then possible to calculate the Charpy-V notch impact requirement to be put on steels and welding procedures for a certain acceptable flaw size.

### 1.13 Discussion

It will be valuable to get better understanding of the way in which different welding parameters influence the properties in the weld metal of Q & T steels. Then it will be possible to achieve joints with mechanical properties matching those of the base material and at the same time to be free from flaws.

The properties can be affected in three simple ways, in negative sense. The first case is when the arc energy input and/or preheat temperature is high and outside the “parameter window”. The second is, when the arc energy input and the preheat temperature is low and also outside the “window”. The third is when the welds are affected by a temperature cycle similar to a thermal stress relieve where some elements segregate into the grain boundaries and increase the risk for a drop in toughness. Phosphorus is known to segregate to the grain boundaries at temperatures between 450 and 550 °C.

This “window” for the arc energy input is getting smaller as the strength of the base material is increased. It is at the same time more difficult to achieve matching strength i.e. 0.2% proof strength in the weld metal, as the strength of the base material is increased.

When welding in the vertical up position with manual metal arc welding, weaving the electrode is a common practise. A restriction often used is to restrict the weaving width to three times the electrode diameter. Weaving wider affects the arc energy input to be higher and the run out length to be shorter. In those cases where this limitation is not followed a degradation in toughness can be noticed and an accentuated drop in strength and toughness in high strength weld metal, due to the risk of being outside the small “window” is greater.

In the case of low arc energy input the risk for hydrogen cracking appearing as transverse cracks is great especially if the preheat temperature is low. In cases where cold wire or metal powder is added, the weld bead sizes increases and at the same time the diffusion distance increases. The cold wire addition increases the hydrogen content in the weld if the wire has a contaminated layer of dirt from the drawing of the wire. Highly alloyed wires are particularly difficult to make and needs proper lubrication during the manufacturing process. Also metal powder will by its greater surface area increase the hydrogen content.

The transverse cracks are not easy to detect by radiographic inspection. The weld reinforcement has to be removed by grinding so that the ripples on the weld bead should not

hide a defect. If the thickness of the weld is greater than about 40 mm it is not possible to detect these thin transverse cracks. Inspection by ultrasonic testing will require that the weld reinforcement is removed and the surrounding area to be ground flush with the plate. Defects close to the surface are difficult to detect. Sizing of the defects is difficult. It depends among other things on how the defects are oriented in relation to the direction to the beam of sound.

It is not likely that all defects are found. Especially the transverse cracks are difficult to detect in practise. Up to 50% of the defects will not be found during normal routine inspection of a welded joint.

The sizing of these flaws is also difficult especially if they are small.

A question is, can they be accepted and if so then what size can be accepted?

Calculations made on defects with the size 5 mm deep and 10 mm long half elliptical surface defect give regarding fatigue sufficient life for a submarine while other structures will not survive. Submarines may be impact loaded by depth charges and this impact can create a brittle fracture if the weld and plate is not tough enough.

Calculations made show the level of requirements for impact properties to be put on the welding procedure to avoid brittle failure for a certain flaw size.

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## 2.1 SUMMARY OF THE PAPERS

### 2.1.1 Paper A:

#### **Effects of Welding Parameters on Weld Zone Toughness and Hardness in 690 MPa Steel**

Specifications for the welding of high-strength steels are generally intended to control hydrogen cracking and provide adequate weld zone toughness for resistance to fatigue cracking and shock loading. The specifications should also allow welding to be undertaken safely and profitably.

The work described here was designed to identify the optimum choice of welding parameters, notably preheat, interpass temperature and heat input, for the welding of a 690 MPa (100 ksi) micro alloyed quenched and tempered steel.

The paper covers investigations into two aspects of weldability: toughness and hardness. The first part involved shielded metal arc (SMA) butt joint welding of carefully designed plates at a range of preheat and interpass temperatures and heat input values, to identify welding procedures that give maximum HAZ and weld metal toughness.

The test procedure for the first investigation involved SMA welding at preheat and interpass (P and I) temperatures from  $-20^{\circ}\text{C}$  to  $220^{\circ}\text{C}$  using heat input values of (approximately) 1.3, 2.9 and 4 kJ/mm.

Charpy V-notch energy and fracture appearance transition curves were then generated with the Charpy notches being carefully located to give the lowest toughness values.

Results showed that the minimum preheat and interpass temperature for control of weld metal hydrogen cracking was  $60^{\circ}\text{C}$ . These low values of toughness in both weld and heat affected zone consistently occurred in the weld root region. Low preheat techniques and one-sided welding should therefore be avoided for critical applications.

For weld metal, greatest toughness occurred at high preheat and interpass temperatures ( $160^{\circ}\text{C}$ ) combined with low welding heat input (1.2 kJ/mm). For high heat input welding, preheat and interpass temperature had little influence on toughness over the range of temperatures examined.

For heat affected zone regions, levels of toughness obtained when using the high preheat/low heat input and low preheat/high heat input techniques were similar. It was found, however, that careful control of preheat and interpass temperature was essential, for each heat input value, because toughness can drop off rapidly on either side of the optimum interpass temperature. For some applications, optimum preheat and interpass temperatures were found to lie between  $60^{\circ}\text{C}$  and  $90^{\circ}\text{C}$ .

In the second study, a series of bead-on-plate submerged arc welds was deposited using a wide range of preheat and interpass temperatures, voltages, currents and travel speeds. In each case, the welding parameters were chosen so that the welding power input ( $V \times I$ ) could be held constant while varying heat input, or vice versa. For each weld, hardness readings were taken in both the weld deposit and HAZ at approximately 1 mm from the weld interface.

It was found that:

- 1) HAZ hardness readings (390-430 HV30 for heat input values up to 2 kJ/mm) were consistently higher than weld metal hardness readings.

- 2) Heat input has a marginal effect on HAZ hardness up to about 2 kJ/mm, however, above 2kJ/mm the hardness drops off at a rate of approximately 30 HV30 for each increase of 1kJ/mm.
- 3) Weld metal hardness dropped continuously over the range of 0.5 to 4.5 kJ/mm.
- 4) Both HAZ and weld metal hardness drops approximately 1 HV30 for every 4°C increase in preheat and interpass temperature.

The results of this work have been used to develop an optimised weld design for butt joint welding of 35 to 50 mm plate. This uses high heat input for all but the weld capping passes, where high interpass temperatures and low heat input techniques are employed.

### **2.1.2 Paper B:**

#### **Submerged Arc Welding with Alloy Powder Additions for High Strength Steels**

As part of an investigation into improving productivity in fabrication of high stress steels, a 50 mm thick, 690 MPa quenched and tempered steel was welded by the submerged arc process with alloyed metal powder additions. The amount of metal powder added was 0, 2, 4, 6, and 8 kg of powder per hour. With additions of 8 kg/h the deposition rate was approximately double that obtained without the powder additions.

Despite using the standard techniques for control of hydrogen cracking, transverse cracks were in most of the welds containing powder additions. The amount of cracking was increased as the amount of metal powder was increased. The cracking was identified as hydrogen induced cold cracking (HICC) by metallographic techniques.

During mechanical testing of the welds, all tensile tests (except one) failed at pre-existing cracks and the tensile test results were consequently well below specification requirements. The Charpy tests, however, were not adversely affected by the existence of cracking, illustrating that the weld metal in all cases was inherently tough.

An explanation for the cracking problem in this multipass weld is discussed. In this investigation the preheat and interpass temperature was kept between 60 to 120 °C, the test pieces were small and the time between passes was short. The diffusion rate for the hydrogen is relatively low at these temperatures. It is believed that cracking is due to a build up of hydrogen in successive weld passes as the rate of diffusion of hydrogen is unable to keep up with the rate at which hydrogen is fed into the weld. Since the size of the weld bead increases as the amount of added metal powder is increased, the welds with increasing alloy powder content have an increasing sensitivity to cracking.

### **2.1.3 Paper C:**

#### **The influence of wide weld beads on the mechanical properties for manual metal arc welded high performance steel**

In order to investigate the effects of different weave pattern on the mechanical properties, two welded tests were done with two different electrodes, Atom Arc 10018 M1 and Atom Arc 12018 M2.

The heat input value increases as the run out length get shorter. A wide weld bead is normally regarded to be welded with high heat input.

One side of the test plate was welded by a weave pattern to simulate a bulky triangular weave.

On the other side of the test plate the weld was welded by a wide weave pattern greater than allowed in the qualified welding procedure.

The investigation shows that there is a clear drop in toughness for welds made by the electrode Atom Arc 12018M2 and the weld performed with a bulky weave technique.

The electrode Atom Arc 10018M1 did not show this drop in toughness.

Th toughness in the heat affected zone was not greatly affected in these tests.

#### **2.1.4 Paper D:**

##### **Effect of Post Weld stress Relief Heat Treatment on Charpy Impact Toughness of 690 MPa Yield Stress Weld deposits**

A program of Charpy-V notch testing has been undertaken to study the effect of stress relief heat treatment and ageing on high strength steels and weld deposits. Duplicate tests were undertaken in Sweden and Australia.

The results showed that such treatments had little effect on two grades of 690 MPa yield stress quenched and tempered parent metals and on a 550 MPa cast steel. Ductile/brittle transition for these metals was in the range of -50°C to -80°C.

However, weld deposits were found to undergo a significant degradation in toughness as the result of post weld stress relief heat treatment at 560°C for two hours followed by air cooling.

Both manual metal arc and submerged arc weld deposits in these steels showed ductile/brittle transition temperatures as high as -20°C to 0°C.

Furthermore, the toughness of weld deposits was found to be dependent on the thermal treatments used on the steel immediately after the post weld heat treatment. The deposits were found to increase in toughness if faster cooling rates was used and weld metal toughness could be fully restored by soaking the welds at 625°C after post weld heat treatment, and subsequently quenching in water.

The effect of ageing these weld deposits at 250°C for up to 1 000 hours was also investigated. For welds that had undergone conventional post weld heat treatment there was no measurable change in toughness. For welds without post weld heat treatment there was evidence that ageing had a slow but detrimental effect on toughness.

#### **2.1.5 Paper E:**

##### **Personal reflections on 20 years of R&D in shipbuilding – its implementation and future aspects.**

This paper present some R&D projects intended in improving quality and productivity while reducing the costs for building ships. My personal views on the implementation of results from different R&D projects into the shipbuilding industry are described with reference to the following examples:

Development of more sophisticated robotic welding systems including sensors will take place, better and cheaper computers will make this happen. The programming of the robots will get better and faster and the programming can be done from the CAD system which may limit the use of drawings. The tolerances for every component will be better and this will mean that the precision in every step will be higher.

The development and use of steels with higher strength will continue. The weld metal is the weak point now, further development is needed to get balance in properties between the weld metal, HAZ and the base material. Life cycle costs can be effected by the choice of material for the ship.

Increased productivity is the goal and this can be achieved by smaller joints to fill and less deformations in laser beam welding, but also with selection of the best welding position were welding methods can be used which have higher deposition rate compared to position welding.

New design can eliminate the need for fine tolerances and then alternative welding methods can be used. New design can also reduce the weight of the components and give lower transportation costs.

#### **2.1.6 Paper F:**

##### **Development of Hydrogen Cracks in High Strength Submerged Arc Weld Metal-An Investigation by Factorial Analysis**

A test for hydrogen induced cold cracking in multipass submerged arc welded joints in high strength steels has been developed, and this paper describes a technique to evaluate different factors which may have an influence on the creation of transverse cracks in the highly alloyed weld metal.

The evaluation of factors effecting the risk for hydrogen induced cold cracking was done by reduced factorial test. Six factors on two levels, high and low, were investigated by a reduced test plan, which resulted in 16 test pieces to be welded. Following factors were investigated: welding current, welding voltage, travel speed, preheat, handling of welding flux and the addition of iron powder to the weld metal.

Preheat and high welding voltage showed a positive effect but addition of iron powder showed a negative effect on the time to get hydrogen induced cracking.

#### **2.1.7 Paper G:**

##### **Estimation of fracture toughness according to Sanz's and Wallin's model, in electrogas welds of about 25 year old ships**

During years many ships have split in two parts, this has focussed the interest on the fracture safety of these vessels. Some have polluted the coast and effected the natural life due to the oil spill and this is some of the less desirable consequences. The question aroused is whether these shipwrecks are due to insufficient materials and welding



technology especially in the early 70-ties

This is discussed, and it is concluded that in terms of the state of materials and welding technological knowledge at the time the ships were built, the designers from the shipyards and the specialists from the classification societies can not be blamed for the disasters.

By looking back at data available at that time and use that data in methods available today it is possible to correlate and check the models given by Sanz and Wallin and the real situation of today.

In the early 70-ties it was discussed how to use the COD value for fracture safe design but the algebraic "design curve" as it is now called did not appear until 1980.

A Japanese report from 1976 could if read the correct way, actually have constituted the definite turning point. In an IIW document, it was shown that a crack nucleated in the weld will continue propagating in the weld if that weld has a poor notch toughness.

Among reports that might have been interpreted as early warnings was a paper from Sweden published at an international symposium 1975 in Osaka. This report gave the Charpy-V notch toughness as also CTOD values of typical electrogas welds as well as those of other high energy welds.

When the above mentioned key reports were published however, many of the shipyards had already closed down the production of very large crude oil tankers from other reasons . After the oil crisis the rise in oil consumption levelled off and actually there was no market for new ships. Many ships that were made went directly to fjords in Norway, were they were laid up. Research projects regarding development of steels, which could be welded by high heat input was started during these years.

In the present paper the significance of some of the early documents is emphasised and some fracture mechanics calculations are performed showing that large defects has to be present in the "right" area simultaneously together with low toughness in the welded joint if the probability for failure should be high.

Corrosion is a factor constituting an added risk, by causing the nucleation sites for premature propagating fatigue, which can grow to a large defect during hard weather conditions.

### **2.1.8 Paper H:**

#### **Estimation of strain rate at explosive loading by Wallin's model and data from Explosion Bulge Test, Flaw Bulge Test and Charpy-V Notch Test to calculate flaw size and impact requirements**

Wallin's model have been used to calculate the impact requirements to be used on welds in extra high strength steels for a certain flaw size.

This impact energy requirement depends on the acceptable flaw size and the yield strength of the material.

The calculation have used values which were obtained from 7 different test programmes involving Explosion Crack Starter Test (ECST) and one test programme involving Flaw Bulge Explosion (FBE). One step in the calculation was to get a figure on the strain rate, which will act on the weld during testing. By choosing a strain rate which will give almost the same impact value as the tested weld show, makes it possible to calculate the strain rate over the weld. As a flaw size for the ECST was the dimensions for the the brittle weld bead welded on top of the weld

The strain rate for ECST is calculated to be about  $10^6 \text{ s}^{-1}$ .

The flaw size in FBE was the size of the defect made by fatigue. The calculated strain rate for the FBE test, which is performed under water, is calculated to be approximately around  $10^4 \text{ s}^{-1}$ .

### **2.1.9 Paper I**

#### **Submerged arc Welding of WELDOX 900E in 50 mm Plate Thickness –Hydrogen cracking tests**

Techniques for submerged arc welding of 900MPa yield stress steel, WELDOX 900E, have been investigated using a new technique for evaluating the risk of hydrogen cracking in multi-pass welds.

Welding procedures investigated included the use of solid wire filler material, solid wire with cold wire additions and the use of a newly developed tubular wire with, and without the addition of high-alloyed cold wire. Preheat and inter pass temperature of 80, 140 and 200 °C were used.

None of the test welds reached the required minimum yield strength of 900 MPa . However, all test welds show relative good impact properties at –40 °C. Welds with the lowest carbon equivalent show the highest impact values; about 100 J at –40 °C.

A new technique for evaluating the risk for hydrogen cracking showed that only one test out of nine did not have cracks. The sample with the lowest carbon equivalent and the highest preheat temperature remained free from cracks during the 114 hours that the test was running. The lowest preheat temperature, produced welds with extensive hydrogen cracking in form of sub-surface transverse cracks. The size of these defects was about 3 by 5 mm.

The hydrogen cracking test method developed showed a good correlation to the SS-EN 1011-2 standard method B for calculation of preheat temperatures. The commonly applied time restriction of 72 hours between finished welding and non-destructive testing is shown to be of importance due to the long crack propagation time found in this investigation.

### **2.1.10 Paper J**

#### **Submerged arc Welding of WELDOX 900E in 50 mm Plate Thickness –Cold Wire Addition and mechanical Properties**

The mechanical properties for submerged arc welding of 900MPa yield stress steel, WELDOX 900E, have been investigated using a new technique for addition of cold wire to the weld in order to get matching or over matching properties in the weld metal.

Welding procedures investigated included the use of solid wire filler material, solid wire with cold wire additions and the use of a newly developed tubular wire with, and without,

the addition of high alloyed cold wire. Preheat and inter-pass temperatures of 80, 140 and 200 °C were used.

None of the test welds reached the required minimum yield strength of 900 MPa. However, all test welds show relative good impact properties at -40 °C. Welds with the lowest carbon equivalent show the highest impact values; about 100 J at -40 °C.

The effect of interpass temperature on the impact properties was small compared to the effect of carbon equivalent. The addition of cold wire has a tendency to improve the impact properties in the HAZ.

The amount of martensite in the microstructure of the weld is decreasing as the preheat temperature is increased. There is an indication that the hydrogen cracking in the investigated weld metals is not related to the martensitic microstructure alone.

## 2.2 CONCLUSIONS

- To obtain optimum properties in a welded joint in high strength steels there is correlation between preheat temperature and the arc energy input.
- This so called window of acceptable properties will be smaller as the strength level rises.
- Optimum toughness of manual metal arc welds in 50 mm high strength steels were obtained by using low arc energy input and a preheat temperature of 180 °C.
- Optimum HAZ toughness in these steels was obtained at both high and low welding heat inputs under stated conditions; however, the appropriate preheat and interpass temperature must be maintained throughout welding. For maximum productivity and improved operator comfort, it is recommended that high heat input welding be used in conjunction with preheat and interpass temperatures close to 70 °C.
- Welding high strength steels and to get matching tensile properties involves higher alloyed filler material compared to the plate material.
- Higher alloyed filler material means that the composition of the filler material decides the level of preheat. The weld will due to higher carbon equivalent be more prone to cracking than the base material.
- The defects will in most cases be small transverse cracks about 3 to 5 mm in size and most of them are buried in the weld and not coming out to the surface.
- Not all of these defects are found by non-destructive testing.
- Requirements for allowable strength and impact properties must take these defects into account and give an acceptable size for a allowable flaw.

## 2.3 FURTHER WORK

In order to find the optimum welding process parameters for different steels with even higher strength than the ones investigated, it is necessary to do more research.

Further research will be needed in following areas:

- Development and validation of testing methods for hydrogen cracking in multipass submerged arc welded joints.
- Development of a model to simulate the effect of hydrogen on the integrity of steel welds.
- Development of consumables giving matching strength in the weld metal for the strength level of 900MPa and higher in 0.2% proof strength..
- Development of productive welding methods for the strength level of steels above and including 900 MPa in 0.2% proof strength.
- Development of the non-destructive testing methods to improve the technique for sizing of detected flaws.
- Validate the techniques for calculations of predicted lifetime of structures made by these high strength steels.