

SAFSS

## STRUCTURAL APPLICATIONS OF

# FERRITIC STAINLESS STEELS

(SAFSS)

Report To:RFCSDocument:Study of weldabilityVersion:02Date:March 2014

Outokumpu Stainless Oy

Terästie, FI-95450 Tornio, Finland Tel.: +358 16 4521, Fax: +358 16 452 620, www.outokumpu.com Domicile: Tornio, Finland. Business ID 0823315-9, VAT FI08233159



Version	Date of	Purpose	Authors	Technical	Approved
	Issue			Reviewer	
01	5.7.2011	Issue to RFCS and TGS8	Sev, HPH	JSk	MYI
02	31.3.2014	Editorial correction	Sev	JSk	MYI

The testing, assessment, findings and conclusions outlined in this report have been made with the intent of due diligence, care and best effort. Despite that we may not be held liable for any loss or damage, either direct, compensatory or consequential, exceeding the amount paid for this report.



#### **EUROPEAN COMMISSION**

## Research Programme of The Research Fund for Coal and Steel - Steel RTD

Title of Research Project:	Structural Application of Ferritic Stainless Steels (SAFSS)
Executive Committee:	TGS8
Contract:	RFSR-CT-2010-00026
Commencement Date:	July 01, 2010
Completion Date:	June 30, 2013
Work Package No and Title:	WP5, Welded connections
Draft Final Report:	WP5.1, Study of weldability
Beneficiary:	Outokumpu Stainless Oy 95490 Tornio, Finland
Research Location:	Outokumpu Stainless Oy Tornio Research Centre
	95490 Tornio, Finland
Contact person:	Jukka Säynäjäkangas
Report authors:	Severi Anttila Hannu-Pekka Heikkinen



## Summary

The report deals mainly with weldability characteristics of currently manufactured low and medium chromium ferritic stainless steels.

Ferritic stainless steels are a wide range of chromium-based steels. Ferritics are weldable if certain precautions are taken into account.

Some of these steels exhibit austenite at higher temperatures, which on cooling tends to transform into martensite. Other steels are entirely ferritic all the way to molten metal, which in some cases exacerbates the grain coarsening in the high temperature heat-affected zone. Lower heat inputs are preferred when welding ferritics.

High martensite content, 90...100% in the heat-affected zone is beneficial, toughnesswise for low carbon grades. Grain boundary martensite as in the case with grade 1.4016 (430) deteriorates the toughness and ductility of the welded connection. Austenitic filler metals can salvage some of the toughness in the welded joint but the HAZ remains brittle and postweld heat treatment is recommended for tempering the martensite. This will also improve the corrosion resistance by allowing the chromium back-diffusion to take place to restore the passive layer of the steel surface.

Sensitisation is mainly concentrated to unstabilised steel grades. Nevertheless, low chromium ones like 1.4003 or 1.4512 (409) can suffer from chromium depletion in some situations. Very low heat inputs can restrict the chromium back-diffusion and expose the steel surface to intergranular corrosion.

Hydrogen embrittlement concerns all ferritics and every precaution should be made to prevent hydrogen access to the weld metal.

Second-phase embrittlement phenomena such as 475 °C, sigma- and Laves-phases concern mainly medium- and higher chromium grades. These are not common in as-welded structures but incorrectly selected postweld heat treatments or service temperatures could lead to these phenomena.



## Contents

Summary	
Contents	
1 Introduction	6
2 Ferritic stainless steels in general	7
2.1 Microstructural characteristics	7
2.2 Ductile-to-brittle behaviour	9
2.3 General corrosion properties	11
3 Welding aspects	12
3.1 Welding filler metals	12
3.2 Shielding gases	13
3.3 Post-solidification phase transformations	14
4 Weldability of ferritics in general	16
4.1 Weldability of ferritic steel groups	16
4.2 Heat-affected zone of ferritics	16
4.2.1 Hydrogen induced cracking	16
4.2.2 Grain coarsening	17
4.2.3 Formation of martensite	18
4.3 Embrittlement phenomena	20
4.3.1 4/5 °C embrittlement	20
4.3.2 Sigma phase embrittlement.	21
4.3.3 Laves phase embrittlement	22
4.5.4 Sensitisation	22
4.5.5 High-temperature emonutement	25
4.4 weiding guidelines	24 24
4.4.1 Low near input	2 <del></del> 24
4.4.3 Postweld heat treatments	25
	20
5 Weldability of common ferrific grades	<b>20</b> 26
5.1 Grade 1.4502	20 27
5.3 Grade 1.4016	27 28
5.5 Grade 1.4509	20
5.5 Grade 1.4521	
5.6 Higher chromium grades	30
6 Conclusions	32
References	33



## **1** Introduction

For some time now, ferritic stainless steels have gained growing interest because the absence of nickel gives them more stable and cheaper raw material price in comparison to austenitic stainless steels. Today, modern production facilities enable better-controlled manufacturing processes concerning alloying and impurity levels [1]. Therefore, modern ferritic stainless steels have the capability to replace many austenitic stainless steel grades.

Welding is an essential joining method for fabrication industry. Weldability of ferritic stainless steels is challenging and many embrittlement phenomena decrease the interest to use these materials. Furthermore, many embrittlement phenomena in the base metal tend to appear in the weld metal if welding autogenously or when similar filler metal is used. Typically ferritic stainless steels are welded with austenitic filler metals, which is essentially a precaution to maintain adequate toughness and act as a hydrogen-sink in the weld metal [2].



## **2** Ferritic stainless steels in general

Early ferritics could not compete with austenitic grades because of their modest corrosion resistance and fabrication properties, including welding. Today, application variants of ferritics are commercially competitive and available. [3] Modern ferritics have good characteristics for heat resistance, oxidation and corrosion. In addition, ferritics have certain advantages compared to austenite grades, such as higher thermal conductivity and lower thermal expansion coefficient. These properties improve weldability and fatigue performance in thermal cycles. Stress corrosion cracking, a common problem of austenitics, does not concern ferritics. [4]

Most distinguish difference between ferritic and austenitic stainless steels is the magnetism. Ferritic phase structure is magnetic whereas austenitic is paramagnetic at ambient. Other relevant physical features for ferritics are higher thermal conductivity and moderately low thermal expansion coefficient. In welding, higher thermal conductivity is desired since the heat flow in weld region is not concentrated as it is when welding austenitics. Higher conductivity distributes the heat efficiently and long-term undesired temperature exposure can be avoided. The thermal expansion coefficient of ferritics equal to that in carbon steels, whereas austenitics have much a higher coefficient, which results in buckling in weldments. Therefore, ferritics distort less in thermal treatments. [5]

Comparing ferritics with austenitic stainless steels is common, but in reality, this comparison should also include mild carbon steels. Corrosion in some form is inevitable when using mild carbon steels. Ferritics are corrosion resistant in atmospheric and non-aggressive environments. Therefore, these steels can offer a very good choice between "pricy" austenitic stainless steels and "rusty" carbon steels. Today's environmental values and total lifecycle costs can easily drop the price gap in favour of ferritics. [6]

## 2.1 Microstructural characteristics

Ferritics are a wide range of chromium steels. Chromium and carbon are key alloying elements determining the microstructure and properties of ferritics. Therefore, these steels are usually divided by chromium content to three main groups: low chromium, medium chromium and high chromium steels. This clarifying is a good starting point to introduce ferritics, although modern listing includes other variables such as the use of stabilising elements, e.g. titanium and niobium, or better corrosion resistance with added molybdenum. [7]

Low chromium steels contain 10.5 ... 13.5 % Cr Medium chromium steels contain 14 ... 18 % Cr High chromium steels contain 20 or higher % Cr



Increasing chromium content usually means better corrosion resistance. Low and medium chromium steels are generally suitable for non-aggressive and atmospheric environments whereas high chromium steels are widely used in industry applications. [5]

A good way to discuss ferritics more specifically is to introduce the iron-chromium equilibrium diagram. A constant carbon section of the iron-chromium-carbon ternary equilibrium diagram is shown in Figure 1A. This diagram describes relatively accurately the phase structures, which ferritic stainless steel undergoes at given carbon and nitrogen content at different temperatures. Low chromium ferritics generally fall into outer gamma-loop ( $\delta + \gamma$ ), where the phase structure is dual phased with the majority being ferrite and a substantial amount being austenite. Medium chromium steels can either have a dual phase or a fully ferritic structure as the carbon and nitrogen contents widen the gamma-loop size, as shown in Figure 1B. [7]



Figure 1. (A) Iron-chromium-carbon ternary equilibrium diagram for carbon content of 0.01 % or less [8]. (B) The influence of carbon and nitrogen to the diagram [9].

The dual phase structure gives low chromium ferritics an interesting combination of stainless steels corrosion resistance and carbon steels engineering properties. This is because during cooling cycle the formation of martensite occurs. Due to the low carbon content, the lath martensite formed is relatively soft and ductile, not as brittle as high carbon martensite known as plate martensite. [10] High chromium steels are fully ferritic at all temperatures, i.e. the phase structure stays ferritic all the way to molten metal. These fully ferritic stainless steels have better corrosion resistance due to their higher chromium content, but they suffer increasingly from various embrittlement phenomena.



Due to sensitisation embrittlement phenomenon, ferritics are often stabilised. When adding e.g. titanium and niobium to composition, ferritics can suitably stand for intergranular corrosion, which exposes steel surface to the sensitisation phenomenon. Intergranular corrosion is restricted when free carbon and nitrogen are trapped by these stabilising elements and formation of various, less severe, carbides and nitrides occurs. [7] Titanium and niobium have a higher affinity for carbon than chromium has, i.e. these carbides form readily and earlier than chromium carbides. Therefore, these formed carbides suppress potential chromium carbide precipitation efficiently. [11]

As mentioned, when classifying ferritics by their chromium content the main information relies on iron – chromium equilibrium diagram. By today's standards, this kind of classification is out-dated. Modern ferritics vary with their interstitial content, formation of martensite and stabilisation attributes. International Stainless Steel Forum (ISSF) describes modern "ferritic families" in Figure 2.



## THE 5 GROUPS OF FERRITIC GRADES

Figure 2. ISSF classification of ferritic grades [5].

## 2.2 Ductile-to-brittle behaviour

An essential matter when speaking of ferritics toughness properties is impact toughness. Ferritics cannot offer very good toughness values because of their high notch sensitivity. This sensitivity relies on interstitial content in medium and high chromium steels [7]. Even though modern ferritics no longer have overall detrimental impact toughness, this is still very much the case with as-welded structures. Extreme grain coarsening and possible formation of martensite in heat-affected zone and weld metal may occur.





Figure 3. A) Ductile-to-brittle transition temperature curve [12]. B) Schematic presentation of martensite laths influence to DBTT [9].

Transition temperatures of ferritics are located at moderately high temperatures, from 20 °C to 80 °C, depending on the composition [9]. This behaviour is also highly dependable on section size, as thinner materials behave in more ductile manner and the DBTT decreases. [13] Low chromium unstabilised ferritic-martensitic stainless steels or martensitic steels in general show somewhat different transition curve, since the combination of small and large martensite laths flatten the DBTT curve, as illustrated in Figure 3B [9].

The poor toughness in ferritics is mainly caused by interstitial elements in bcc structure. Pure bcc structure itself is relatively ductile at lower temperatures but even a small amount of interstitials including carbon, oxygen and nitrogen increase the DBTT considerably. These interstitials lock dislocations efficiently and flow stress increases as the Cottrell model predicts. Lattice friction stress increases considerably when temperature decreases. This leads to more brittle fracture behaviour. [14]

It is known that the increase in the DBTT is also related to the increase of the delta-ferrite grain size [9]. This behaviour is notable with smaller grain sizes but with coarse-grained sections, e.g. high temperature heat-affected zone (HTHAZ), the effect diminishes as shown in Figure 4. Ohashi et al. [15] observed that even though coarse grains promote crack initiation, but crack propagation is only slightly dependent on the grain size. In addition, ferritics are cooling-rate sensitive. Carbide and nitride precipitations either align along the grain boundaries or distribute randomly inside grains. Thus, grain size is an important factor when speaking of ductility and toughness. Finer grains provide greater impact toughness. [7]





Figure 4. Grain size effect on DBTT curves for high chromium ferritics [16].

Other influencing factors are precipitates, e.g. titanium and niobium carbides and nitrides [13]. Precipitates containing titanium act harmfully as crack nucleation sites in a comparatively hard dual phase structure [9]. It is known that titanium additions beyond 0.17 % decrease ductility and fracture stress [17]. Niobium is found to improve toughness more than titanium due to preferable inclusion structure [2].

#### 2.3 General corrosion properties

There are many types of corrosion in stainless steels. Uniform corrosion is caused by chemical or electrochemical reaction between the metal and the environment. [18] Local corrosion can be divided into mechanical and non-mechanical corrosion. Former types of corrosion are stress corrosion cracking, thermal fatigue and erosion corrosion. Latter types are galvanic, intergranular, pitting and crevice corrosion. [2] Ferritics are not susceptible to stress corrosion cracking, which is a common problem with austenitic stainless steels [19]. The thermal fatigue properties are better in ferritics due to the lower thermal expansion coefficient and higher thermal conductivity [4]. However, intergranular corrosion and many other embrittlement phenomena exist in ferritics.



## **3** Welding aspects

Welding – as the process of joining materials using heat or other source to make a permanent bond – is undoubtedly one of the most important process of current fabrication industry. Welding and welding related technology affects substantially nations' income and welfare. Due to this matter, welding of stainless steels must be as effortless as possible. During the years, welding technology has improved as well as steel grades. Welding techniques have generally stayed the same but welding methods have improved steadily, e.g. combined welding processes have been developed. [20]

#### 3.1 Welding filler metals

Successful weld metal can be produced in several ways. A filler metal is used to ensure that the weld metal has optimum corrosion performance and mechanical properties. Welding arc and segregation induce some loss of alloying elements and that is why filler metal compositions have typically slightly higher alloying contents than similar base metal has. [21]

Welding autogenously means without any added filler metal. Weld metal composition is determined entirely by the base metal composition. Typically, pipes and thin materials can be welded this way. Solidification and weld metal nucleation occurs by epitaxial mechanism. That means a base metal grain structure is a substrate to the weld metal microstructure. In ferritics, a HTHAZ is in concern because coarse-grained HAZ may lead to a coarse columnar grain structure in the weld metal. Autogenous welding is not popular with stainless steels as it leads to poor corrosion resistance, loss of ductility and toughness [13]. In addition, crack-susceptible compositions are hazardous to be welded autogenously. [11]

When welding a base metal by a matching filler metal, the formed weld metal microstructure resembles the base metal HAZ to some extent. This is because compositional-wise the base metal and the filler metal do not significantly deviate from each other. Solidification and nucleation happens preferably in an epitaxial manner. [22] As mentioned, filler metal compositions are typically a little bit higher than the corresponding base metal due to various losses and segregation of the elements during welding process. Matching filler metals are used when weldments have to match the base metal corrosion or strength characteristics. This is mainly the case when using ferritic filler metals with similar base metal. Other possible reasons are similar magnetic properties or surface colour [23].

Most used filler metals with ferritics and stainless steels in general have been the austenitic stainless steel filler metals. Their major advantage over ferritic filler metals is the superior toughness properties that predominantly austenitic weld metals provide. Usage and knowledge of austenitic filler metals have been vast. Various compositions



have been introduced over the years to enhance weldability of various metals. In addition, standardisation of austenitic filler metals is wide-ranging. Even though austenitic filler wires are largely preferred when welding ferritics, they are not suitable for sulphuric environments because nickel tends to react with sulphur. High thermal expansion coefficient will also cause problems in thermal-cycling conditions. [13]

Other possible filler wires used with ferritics is high-nickel and duplex-filler metals. High-nickel filler metals suit especially for highly corrosive environments. Their thermal expansion coefficients are similar to those of ferritics, which is an advantage. However, nickel could cause problems with high molybdenum ferritics when formation of brittle intermetallic phases is possible [2]. Duplex (austenitic-ferritic) filler wires are sometimes used as they offer higher yield strength in comparison to austenitic stainless steel fillers [24].

#### 3.2 Shielding gases

Shielding gas surrounds and protects the weld seam. When welding stainless steels, most common shielding gases are argon and helium based mixtures. Helium is an inert gas, which raises the arc energy, improves fluidity and arc stability, and therefore, leads to greater welding speeds [21]. Argon is a better choice to produce a stable welding arc. Adding  $O_2$  or  $CO_2$  prohibits possible undercutting at fusion boundaries. Carbon dioxide increases penetration, but, at high voltages it leads to high level of spatter and increasing carbon content in the weldment. [2]

Typically, argon and helium mixtures are used with TIG process. Only high-purity gases are suitable, as this will decrease the risk for contamination.  $1...2 \% O_2$  or  $2...3 \% CO_2$  are normally added for MAG welding gases to stabilise the arc and to maintain planar metal transfer. However, modern low interstitial ferritic grades could suffer from these oxidising elements. [13]

Main trouble with oxygen in stainless steels is the oxidation that forms on the weld joint as the excess oxygen produces greater formation of slag. Another matter with oxygen is scaling loss. Certain amounts of alloying elements (i.e. Si, Mn, Cr, Nb) will react with free oxygen and form slag. The deficit of these elements naturally reflects to the weld metal composition. [2] This loss of alloying elements is one of the reasons why filler metals have typically higher alloying contents compared to that of the base metals.

If using carbon dioxide as a shielding gas, it dissolves during welding process and the weld metal carbon content will increase by some amount. Avesta Welding [25] and Grönlund [26] studied this *carbon pick-up* from shielding gases containing carbon dioxide, and results are in Figure 5A and 5B, respectively. As can be seen, the weld metal carbon content nearly doubles with 2 % CO<sub>2</sub> shielding gas.







Figure 5. Carbon pick-up reaction in austenitic weld metals: modified from Avesta Welding (A) [25] and Grönlund (B) [26].

Carbon dioxide content should be kept below 3 % to ensure acceptable corrosion resistance. [2] However, some manufacturers prefer higher carbon dioxide contents such as Ar + 8...10 % CO<sub>2</sub>. This noticeably higher carbon dioxide content is said to ensure the wetting (fluidity) of the melt [27]. The usage of this high CO<sub>2</sub> content is questionable since as the carbon content increases, e.g. from 0.01 % to 0.03 % and more stabilisers will be required to ensure the fully ferritic structure. Carbon contents of 0.04C (+0.03N) could expand the gamma-loop up to 20 % Cr. [7]

When welding ferritics any excess hydrogen should not be used. This derives from the cubic crystal structures of ferrite that cannot dissolute solute hydrogen as much as austenite can. Hydrogen induced cracking is discussed shortly in detail. In some specific situations, some amount of nitrogen can be added to shielding gas. Nitrogen is an austenite stabiliser and in certain cases can refine weld metal microstructure. This effect was well documented by Inui et al. [28]. However, excess nitrogen should be avoided as it tends to form pores or react with alloying elements to form various nitrides [2].

#### 3.3 Post-solidification phase transformations

Delta-ferrite phase is the primary solidification product for ferritics [7]. During the cooling stage of a ferritic stainless steel, the amount of austenite increases if austenite phase is stabilised by the composition, i.e. gamma-loop exists. Kaltenhauser has introduced a ferrite factor to estimate welded structures in ferritics. Originally, this equation was derived from a chromium-equivalent, which estimated the solidification behaviour of ferritic base metals. Kaltenhauser's ferrite factor elaborated the martensite content more specifically. The equation is as follows:

$$FF[wt\%] = Cr + 6Si + 8Ti + 4Mo + 2Al - 40(C+N) - 2Mn - 4Ni$$
(1)

When the factor is below 8, the welded structure is fully martensitic. At values 8 ... 13.5 the microstructure is ferritic-martensitic and above 13.5 fully ferritic [29]. Ferrite factor gives a good estimate of the weld metal, but it is not accurate to determine fully ferritic structures. [7] Panton–Kent [30] modified the ferrite factor by adding the



effect of niobium. The studies were made for 9Cr1Mo steel grades, but it is suitable for other steel grades also. [31] The equation is as follows:

$$FF[wt\%] = Cr + 6Si + 8Ti + 4Mo + 2Al + 4Nb - 40(C+N) - 2Mn - 4Ni$$
 (2)

The Schaeffler diagram is a constitution diagram suitable for stainless steels weldments, for it has been derived from actual weld metal microstructures [2]. Due its broad scale, it is useful for evaluating dissimilar compositions. However, the diagram has some limitations, e.g. titanium and aluminium contents are ignored both of which are frequently present in ferritics and weldments. More recently, Balmforth et al. [32] introduced an improved constitution diagram especially for ferritic stainless steel weldments. This diagram is a detailed segment of the Schaeffler diagrams M+F section. This ferritic-martensitic constitution diagram is shown at Figure 6. The diagram was developed to estimate more properly the content of martensite in welded structures.



Figure 6. A ferritic-martensitic stainless steel constitution diagram [32].



## 4 Weldability of ferritics in general

Weldability is often described as the ease on the welding process and the feasibility of the welds obtained to their designed service. The term includes mechanical and corrosion properties of the steel. For the mechanical standpoint, it is important that strength, ductility and toughness are plausible. For the corrosion standpoint, the stress corrosion cracking, intergranular and overall corrosion protection should be considered.

## 4.1 Weldability of ferritic steel groups

*First generation* ferritics delivered good wrought properties. However, welding of those was problematic since ductility, toughness and corrosion resistance degraded. These moderately high carbon and nitrogen ferritics suffered from intergranular corrosion, grain coarsening and in many instances formation of martensite. Typically welding procedure required both preheating and postweld heat treatment. [13]

Introduced stabilisers, other alloying elements and increased chromium contents exposed the *second generation* ferritics to other embrittlement phenomena. Fully ferritic structures suffered grain coarsening at HAZ and second phase embrittlement phenomena such as sigma phase embrittlement. *Third generation* ferritics are comparable to modern ones. Low and well-controlled interstitial contents and added stabilisers improve the weldability substantially. Formation of martensite is no longer an issue since even the low chromium stabilised steels stay fully ferritic in as-welded structures. Therefore, preheating is not necessary. [13] However, fully ferritic structures are susceptible to grain growth and preheating naturally decelerates the cooling rate. Therefore, preheating should actually be avoided. [7]

## 4.2 Heat-affected zone of ferritics

Due to the broad scale of ferritics, also HAZs vary notably. Generally, the most problematic section for ferritics is the HTHAZ. Grain coarsening, formation of martensite, hydrogen induced cracking and various embrittlement phenomena will be discussed.

## 4.2.1 Hydrogen induced cracking

All ferritic structures are susceptible to hydrogen embrittlement, also called hydrogen induced cracking (HIC). When welding ferritics any excess hydrogen will at cooling tend to diffuse towards the HAZ. Any porous that form during solidification can then be potential sites for hydrogen atoms to gather and form  $H_2$  molecules. When the temperature decreases quickly these molecules are stuck in their place and increasing pressure will eventually lead to cracking. [10] Ferritic microstructure is hazardous for hydrogen, in particular. Ferrite body-centred cubic structure dissolute only small



amounts of hydrogen as the austenite face-centred cubic structure is much more applicable for hydrogen [2].

When welding ferritics, several precautions must be taken to account HIC. Problems are usually resolved with preheating and postweld heat treatments or by using an austenitic filler metal, which will act as a "sink" for containing hydrogen. [33] Nevertheless, any hydrogen access should be eliminated in the weld area. Hydrogen shielding gases should be avoided. Moisture, water vapours and oils are also possible sources for hydrogen. Hydrogen can either cause cracking right after the welding operation or reduce ductility without any visible indication. [13]

Hydrogen outgassing or dehydrogenation treatments will remove the excess hydrogen effectively at temperatures of 90...200 °C. At room temperatures, hydrogen outgassing requires several days to complete. [10] Furthermore, short-term treatments can even-out the hydrogen content within the welded connection and decrease the risk of HIC [34].

#### 4.2.2 Grain coarsening

Grain growth is a significant embrittlement phenomenon for ferritics. Grain coarsening becomes relevant only at high temperatures, over 1100 °C or so, depending on chemical composition. Naturally, the focus is turned to welding or other thermo-mechanical treatments [7]. HTHAZ is especially under concern where grain sizes can easily grow beyond 200  $\mu$ m. However, when temperature is kept below 900 °C, grain growth is limited despite the microstructure [9]. Therefore, as it is evident that high heat input causes grain growth, minimum welding heat input is preferred.

Grain growth embrittlement is closely related to the lack of phase transformation as ferritics are notably ferritic, if not entirely. Body-centred cubic structure allows fast diffusion rate and the grain growth phenomenon is mostly controlled by the diffusion. Low and medium chromium ferritics have some amount of austenite at these elevated temperatures and this content is advantageous to suppress the grain growth phenomenon [9]. To be precise, significant quantities of austenite at elevated temperatures can hold back ferrite grain growth due to boundary-pinning effect [7]. Because grain growth is suppressed when austenite exists in the solid solution, low and medium chromium ferritics are tolerable for higher heat inputs. This reaction is only viable when austenite exists at higher temperatures during heating. Some amount of austenite forms during cooling, but this naturally does not hold back the grain growth as it has already taken place. Unsurprisingly, the lack of phase transformation is the main reason for grain coarsening in the HAZ of fully ferritic stainless steels. Stabilised steel grades can bear this phenomenon slightly better because niobium carbonitrides restrict grain growth by pinning grain boundaries. However, this may not be as effective as previously mentioned austenite. [9]



Ultimately, during welding, some degree of grain growth in the HTHAZ is inevitable and to minimize this risk the lowest possible heat input is preferred. This precaution is obviously most important when welding fully ferritic stainless steel grades.

#### 4.2.3 Formation of martensite

Low and medium chromium ferritics may have some amount of austenite at high temperatures. During the welding thermal cycle, the austenite is favourable since it restricts the grain growth caused by welding. However, during cooling this austenite transforms to hard and possible brittle martensite. The nature and the amount of martensite is mainly determined by the steel composition and the cooling rate. If the cooling rate is rapid enough, it can effectively suppress the transformation of delta ferrite to austenite and eventually to martensite. Slower cooling rate within the dual phase area (gamma-loop) will lead to higher martensite content. Zaayman [9] describes accurately the phase transformations during welding of low chromium ferritics. A typical phase transformation of dual phased ferritic-martensitic stainless steel is as follows:

During heating

I) ferrite transforms partially to austenite at the Ac<sub>1</sub> temperature,

II) austenite starts to transform to delta-ferrite above Ac<sub>4</sub> temperature,

III) whole structure is delta-ferritic after the  $Ac_5$  temperature is reached. If austenite stabilisers expand the gamma-loop significantly,  $Ac_5$  temperature does not exist.

During cooling

IV) delta-ferrite starts to transform partially to austenite below  $Ar_5$  temperature (if exists),

V) austenite will transform into martensite below M<sub>s</sub> temperature.

It should be noted that some compositions can undergo different phase transformations, i.e. fully austenitic structures, but low chromium steels commonly align to dual-phase area and can have the dual phase structure. [9]

As described previously, during welding the phase transformation to austenite occurs. The amount of austenite is determined mainly by the duration of the heating cycle since the cooling cycle is too short for austenite to transform into ferrite below  $Ar_1$  temperature (IV-V). Some amount of delta-ferrite transforms back to austenite during early stages of cooling at higher temperatures (IV). This amount of austenite is greater than the amount, which forms during heating from alpha-ferrite. If the compositions consists more austenite stabilisers, the extended gamma-loop will increase the austenite



proportion. The Ac<sub>5</sub> temperature, where the structure is fully delta-ferritic, may not exist and as a result, solid solution consist some amount of austenite all the way to the liquidus. This will naturally reflect on the amount of martensite, which forms during the cooling cycle. Consequently, final structures are predominately martensitic with islands of delta-ferrite. [9]

Welding heat input naturally effects on the phase balance. The higher the temperature peak is, the longer the delta-ferrite is able to transform to austenite and later during cooling to martensite. Most of the austenite forms during cooling, and it is not able to restrict grain growth during heating. Welding is such a rapid thermal reaction that the phase balance does not match entirely the chemical compositions. That is, a little amount of overheating occurs and transformation temperatures during a heating cycle will always be higher than equilibrium phase diagram indicates. [9]

Carbon plays an important role on what type of martensite is formed. If carbon content is moderately high, then the formed martensite during cooling is "high carbon" plate martensite, which is brittle and not desirable. Low carbon martensite known as lath martensite behaves relatively soft and ductile. Therefore, in low and medium chromium ferritics this type of martensite is favourable. Ferrite dissolves only little amount of carbon and rest of it is gathered around grain boundaries. Several researchers have reported that grain boundary martensite should be avoided at all cost as it weakens toughness properties of the HAZ. However, low carbon martensite is preferred structure within certain limits. Zaayman observed that high martensite content (over 90 %) in the HAZ is beneficial for impact toughness, as illustrated in Figure 7A. [9]

Carbon and nitrogen are strong austenite formers and the higher the austenite stability, the higher the martensite content in the HAZ. Figure 7B demonstrates the influence of carbon and nitrogen on impact toughness in predominantly martensitic structure. Higher interstitial content increases the amount of austenite and restricts the delta-ferrite grain growth, but beyond certain point, the increasing hardness deteriorates toughness. Optimum amount of carbon and nitrogen exists somewhere between these two phenomena. [35] According to Zaayman [9], it is of greater importance to ensure that at least 95 % martensite forms in the HAZ than to reduce the hardness values below a reasonable 300 HV, e.g. with niobium additions.







Figure 7. The influence of martensite content on the DBTT at a constant delta-ferrite grain size on 12 % chromium steel (A). Effect of carbon and nitrogen to DBTT on predominantly martensitic structure (B). [9]

#### 4.3 Embrittlement phenomena

When ferritics are exposed to certain temperatures, various embrittlement phenomena are expected. Main embrittlement mechanisms are secondary phase embrittlement and high-temperature embrittlement (HTE). The latter includes a trendy topic sensitisation, i.e. exposure to intergranular corrosion. Secondary phase embrittlement phenomena only occur after a long exposure in a certain temperature regime. On the other hand, welding or other thermo-mechanical processing treatments can cause these embrittlement phenomena to take place much faster than expected due to segregation. [7]

#### 4.3.1 475 °C embrittlement

When ferritics are exposed for long periods in the temperature regime of 400 to 550 °C the chromium-rich ferrite phase precipitates from ferritic matrix. This second-phase formation affects both the mechanical and the corrosion properties of the steel. The coherent precipitation peaks at 475 °C and therefore it is called 475 °C embrittlement. This reaction is evident for all ferritics from chromium content of 12 to 70 %. Roughly above 17 % chromium steels this precipitation takes place without initial threshold and phase structure is usually more brittle. Therefore, this phenomenon is more relevant to medium and higher chromium ferritics. [36] Figure 8A illustrates how 475 °C embrittlement affects the hardness profile of high chromium ferritic stainless steel grade 1.4521 (444). Eventually, this embrittlement phenomenon causes a drop of impact toughness, as Figure 8B presents. Fortunately, an annealing treatment at 570 °C to 675 °C restores the steel properties within 60 minutes [37].





Figure 8. Effect of 475 °C embrittlement on grade 1.4521 (444) ferritic stainless steel. Increase of hardness profile (A) and decrease of impact toughness (unnotched specimens) (B). [38]

#### 4.3.2 Sigma phase embrittlement

Sigma phase is a general term to describe intermetallic phases that form during a long exposure at various temperatures (600 °C to 800 °C). In stainless steels, these phases contain mostly iron, chromium, molybdenum and silicon. The phases are extremely hard and brittle due to their complex tetragonal structure. Main cause for sigma phase formation in ferritic structures is the atomic diffusion rate. In ferritic structures, the diffusion rates are up to hundred times faster than in austenite. This is due to the bcc structure, which is more spacious. Austenite is much more tightly packed; it does not allow atomic diffusion that easily. The peak of these diffusion rates are at 700...800 °C. [2] Sigma phase precipitates at the grain boundaries and gradually transforms the whole microstructure as presented in Figure 9. The formed sigma phase is clearly visible after 150 hours (B), and after 300 hours the whole phase structure has transformed (C). Measured sigma phase hardness can go up to 700 HV as the adjacent ferrite matrix is only 250 HV [39]. Naturally, this difference reflects also to the general hardness profile.



Figure 9. Formation of sigma phase at high chromium ferritic AISI 447. [39]

Even though ferritics below 20 % chromium content are not prone to sigma phase embrittlement it can be considered when welding. The alloying elements more or less



segregate depending on solidification structures and these elements can form harmful intermetallic phases. Therefore, this reaction may occur in weldments even if the base metal is not particularly susceptible to sigma phase embrittlement. [40]

#### 4.3.3 Laves phase embrittlement

Other fairly similar embrittlement phenomenon as sigma phase embrittlement is the Laves phase embrittlement, which concerns high chromium, molybdenum containing and/or stabilised ferritics. The structure of Laves phases varies depending on composition; e.g.  $Fe_2Nb$ ,  $Fe_2Mo$  and  $Fe_2Ti$  have been appointed as Laves phases. Molybdenum content is found to increase the risk of Laves phase embrittlement. Example of Laves phase embrittlement is shown at Figure 10.



Figure 10. Laves phase embrittlement at grain boundaries of molybdenum rich grade 1.4521, annealed for 200 hours (A) and 470 hours (B) at 750...770 °C [37].

Laves phase forms around grain boundaries and dislocations. Laves phase has two independent nucleation mechanisms. At lower temperatures, 600...675 °C, the nucleation occurs on dislocations. At higher temperatures, 750...825 °C the grain boundary nucleation becomes dominant. The Laves phase decreases the high temperature strength, but other possible effects of niobium rich precipitates are being disputed. [41]

#### 4.3.4 Sensitisation

Stainless steels rely their corrosion resistance on dense chromium oxide layer that passivates the surface of the steel. In *sensitisation* this chromium oxide layer is not properly formed. Chances are that significant amount of chromium has reacted with the free carbon that was packed at the grain boundaries in the solid solution. This results in chromium rich carbides, usually  $Cr_{23}C_3$  or comparable nitrides. As seen from the chemical composition, these precipitates hog a great amount of chromium and the passive layer becomes unstable especially in low chromium steels. When this reaction



happens, steel becomes susceptible to intergranular corrosion and this exposure risk is called sensitisation [8]. In Figure 11 are represented the chromium depleted zones in austenitic stainless steel (A) and samples of ferritic stainless steels (B).



Figure 11. Sensitisation in grade 1.4301 (304) austenitic stainless steel (A) [11]. Similar reaction in 1.4512 (409) ferritic stainless steel samples (B) [37].

The origin behind this reaction is a temperature regime (under 700 °C), where the chromium diffusion is low but chromium can still form carbides [42]. When temperature rises to 760 °C, chromium diffusion has enough energy to balance out chromium-depleted areas and some of the chromium rich carbides dissolve back to the matrix. [11] Modern ferritics have very low and well controlled carbon contents. Even though ferrite dissolves very small amount of carbon, there is not free carbon left to react with chromium to form carbides. This is not always the case and other alloying components are introduced. Titanium and niobium can react with carbon and nitrogen to form various carbonitrides. [42] This reaction occurs during manufacturing process or solidifying weld metal and carbides formed are very stable. This process is called stabilisation, as mentioned.

#### 4.3.5 High-temperature embrittlement

Medium and high chromium ferritics, which stay fully ferritic at all temperatures are susceptible to high-temperature embrittlement, also known as HTE. This reaction is closely related with grain coarsening phenomena and sensitisation. It only occurs during thermo-mechanical treatments or welding.

At high temperatures, carbon and nitrogen are dissolved in the solid solution along with the delta-ferrite matrix. During cooling ferrite quickly saturates with interstitial elements of carbon and nitrogen. The excess carbon and nitrogen atoms can react with chromium to form chromium carbides or nitrides. This precipitation happens inside the grains as well as at the grain boundaries. Inner carbides increase hardness and strength and grain boundary carbides expose steel to intergranular corrosion. The high Cr means



more excess interstitial elements exist, since the amount of iron is reduced and the solubility of these interstitials diminishes. Therefore, interstitial content should be kept extremely low (under 250 ppm) in high chromium ferritics. [7]

High temperature embrittlement diminishes by using stabilised steel grades. Titanium and niobium form relatively stable compounds and reduce the excess carbon and nitrogen levels. Even when the HTE has happened, toughness and ductility can be restored by annealing at 800 °C while the formed precipitates overage. [7]

## 4.4 Welding guidelines

#### 4.4.1 Low heat input

Typical solution to prevent large grain size in the weld metal and in the HAZ is to minimize the welding heat input. Low heat input will also increase the cooling rate in the weldment and the martensite content will diminish. [29] Naturally, the selection of welding technique is crucial. Arc welding process is not an ideal technique because of the high heat input and low power density. Therefore, the target should be high energy beam welding processes such as electron beam welding or laser beam welding [43]. However, usability and costs of these processes are demanding. In TIG and MAG welding, the usage of pulsed metal transfer decreases the heat input and this preferably restricts the excess grain growth.

## 4.4.2 Preheating

Preheating decreases the temperature gradient between the weld area and the unaffected base metal. It is done to assure the dehydrogenation and to mitigate the residual stresses of the welded structure, e.g. martensite. This is reasonable action with thicker sections, but with sheets, preheating temperature does not prevent martensite transformation. Nor preheating affects majorly post solidification martensite tempering. With austenitic and martensitic stainless steels the preheating temperature is set to 100 °C [2] but with ferritics preheating temperatures near 250 °C have been used [33]. ASM guides that sections sizes below 6 mm would not need preheating but considerations should be made with joint design, restraint, welding method, cooling rate and dissimilar compositions. [13]

When using ferritic filler metals, hydrogen induced cracking is a risk in the weld metal as well. Ferritics suffer from hydrogen induced cracking and therefore it is mandatory to prevent any hydrogen accessing the weld process. Nevertheless, some amount of hydrogen will still absorb from the surroundings, e.g. slag, air, filler metal. Weld area residual stresses will be decreased when using preheating. In addition, the temperature



is high enough to prefer ductile behaviour of the base metal i.e. when compositions, which are brittle at room temperatures, may behave ductile at 100...200 °C. [43]

Preheating should be avoided with fully ferritic structures since they are susceptible to grain coarsening. Preheating decelerates the cooling to room temperature and allows more time for grain growth to take place in the HAZ. Unnecessary preheating will then enhance this phenomenon as the partially melted coarsened grains in the HTHAZ may nucleate and produce the columnar grains structure in the weld metal. [7]

#### 4.4.3 Postweld heat treatments

When needed, postweld heat treatments (PWHTs) are mainly done for two purposes: for hydrogen outgassing and for tempering of the martensite. Hydrogen outgassing is usually done first, immediately after the welding and ideally so that the welded joint never drops below 100 °C. A proper temperature should be below the  $M_f$  -temperature because after cooling through  $M_f$  -temperature all austenite is transformed into martensite and hydrogen diffusion is more rapid. If any austenite is left, it will reduce the hydrogen removal rate and cause cracking.  $M_s$  and  $M_f$  -temperatures can be calculated from the following equations [2]:

$$M_{s}(^{\circ}C) = 540 - (497 \times \%C + 6.3 \times \%Mn + 36.3 \times \%Ni + 10.8 \times \%Cr + 46.6 \times \%Mo)$$
(3)  
$$M_{f}(^{\circ}C) = M_{s} - 100$$
(4)

With low chromium and low carbon ferritics the  $M_f$ -temperature is between 250 °C and 300 °C, which is relatively high because of the absence of nickel.

Tempering is only carried out for those ferritics, which transform during welding partially to austenite and later cooling to martensite. Therefore, usually, only the first generation ferritics are postweld heat-treated, in particular grade 1.4016 (430). Tempering softens the hard and brittle martensite. Simultaneously, the excess carbon in ferrite will precipitate as carbides. [44] In addition, the internal stresses that welding has caused will decrease. PWHT becomes evident with thicker materials, somewhere 8 mm and above [2].

Recommended heat treatments vary quite a lot depending on the literature source but roughly 750 °C for 1...2 hours is normally suggested. Nevertheless, tempering should be performed above 500 °C to minimize the risk of 475 °C embrittlement. Typically, the cooling procedure includes a furnace cooling which is done to minimize the distortion that can occur with more rapid cooling rates. Furnace cooling should reach 600 °C, followed by air cooling to prevent, again, the 475 °C embrittlement. Usually for thicker sections, air cooling is forced with pressure or water spray. [13]

PWHTs are also effective removing high temperature embrittlement (HTE) and improving overall corrosion resistance. As it is well known, any grain refinement is



impossible to conduct with PWHT for ferritics, as it requires some form of metal working to achieve recrystallisation. [13] Modern high chromium ferritics rarely need PWHTs. This is mainly because of the absence of martensite and, on the other hand, the increasing risk of second phase embrittlement phenomena.

## 5 Weldability of common ferritic grades

## 5.1 Grade 1.4003

Grade 1.4003 is essentially a ferritic-martensitic stainless steel or *semi-ferritic* steel. This low chromium steel grade is equivalent to UNS S40977. It resembles the 3Cr12 steel introduced in the late 1970's but with a lower carbon content of 0.01 %. This steel is to be considered as a link between carbon steels and corrosion resistant alloys, with carbon steel engineering properties and moderate corrosion resistance. The dual phase structure becomes advantageous when speaking about welding properties. In comparison to fully ferritic structures, HAZ grain coarsening during welding is restricted by the formation of austenite. During cooling, the low carbon content of austenite results in a fine lath martensitic structure, which is advantageous for toughness. The extent of martensite is dependent on chemical composition (ferrite factor) and heat input, as the lower heat inputs results in rapid cooling rates and suppress the austenite nucleation [45]. In addition, the predominately-martensitic structure is not normally vulnerable to sensitisation as most of the carbon is dissolved at martensite [21]. These are the main properties that have awakened the recent interest to this steel grade. [6]



Figure 12. HAZ of grade 1.4003. Notice the martensitic substructures: large lath martensitic substructure with potentially retained delta-ferrite (A) and small lath martensitic substructure (B) [34]



Austenitic filler metals are preferred. Martensitic filler wires are also an option, but extra care must be paid to welding conditions e.g. preheating and PWHT (stress relief). Stabilised ferritic filler metals are not ideal because for adequate toughness the welded joint has to be predominantly martensitic. MAG welding is usually done with Ar/He-based shielding gas with some added oxygen or carbon dioxide for arc stability. When using carbon dioxide, carbon pick-up reaction should be considered, as discussed previously.

## 5.2 Grade 1.4512

Grade 1.4512 represents stabilised low chromium ferritic stainless steel, comparable to AISI 409. Due to titanium stabilisation, this steel is fully ferritic. Some compositions may exhibit a small amount of martensite during welding [9]. Stabilisation improves weldability by preventing the sensitisation effect, but recent studies have also indicated that titanium stabilisation alone can lead to the exposure of intergranular corrosion, when excessive high heat input are used and welded joints are then exposed to temperatures of 500...600 °C [46]. On contrary, sensitisation is also found when using low heat inputs. This *Mode 3* sensitisation occurs when excess carbon in ferrite-ferrite grain boundaries reacts with nearby chromium and the fast cooling rate of low heat input restricts the chromium back-diffusion [47]. Titanium also decreases HAZ toughness to some extent by forming brittle precipitates [6]. In addition, titanium precipitates do not effectively pin the grain boundaries during welding, which exacerbates the grain coarsening. [34]



Figure 13. HAZ of grade 1.4512. Notice the vast grain coarsening in HTHAZ. [34]



Austenitic filler metals are generally preferred, like ER308L. Ferritic filler metals e.g. ER409, ER409Nb or ER409LNb are most commonly used with automotive exhaust components where the sour environment and thermal stresses cause failures in austenitic weldments. However, limited toughness is axiom when using ferritic filler metals [34]. MAG-welding is usually done with Ar/He-based shielding gas with some added oxygen for arc stability.

#### 5.3 Grade 1.4016

Steel grade 1.4016 is currently the most used ferritic stainless steel grade in the world, according to the ISSF [5]. This medium chromium steel corresponds to the industry standard of AISI 430. Chromium content of 16 % gives better corrosion resistance and it challenges the dominant austenitic stainless steel grade 1.4301 (AISI 304) in some applications. It is primarily ferritic, but welding will cause some amount of martensite to the HAZ due to higher carbon content and the lack of stabilising elements. Martensite forms typically along solidification and crystallographic boundaries. The limited gamma-loop results in lower austenite content and HTHAZ becomes delta-ferritic during welding cycle. This results in pronounced grain coarsening for the austenite is no longer restricting the grain growth. Therefore, low heat input is mandatory. Absence of stabilisation means that this steel is distinctly vulnerable to sensitisation phenomenon during welding, even though martensite hogs a great amount of carbon and nitrogen. [7]



Figure 14. HAZ of grade 1.4016. Notice the grain boundary martensite in HTHAZ (A) and LTHAZ (B). [34]



Overall, the semi-martensitic HAZ has miserable toughness properties. A PWHT (tempering) is preferred for it softens the hard and brittle grain boundary martensite. Typically this is done around 750 °C for 1...2 hours. Toughness and corrosion resistance is expected to improve. [34]

Austenitic filler metals are preferred and used to salvage some toughness in the welded joint. If using ferritic filler metals, extra low carbon content is required to prevent grain boundary martensite in the weld metal. Titanium stabilised ferritic filler metals are recommended because these will increase the number of equiaxed grains in the weld metal. [34] In some applications, a nickel-based filler metal EN ISO 6082 type is advantageous [21].

## 5.4 Grade 1.4509

Grade 1.4509 is a higher medium chromium grade. This 18 % Cr steel has the best corrosion resistance among the steels previously mentioned. Compositional-wise, it resembles AISI 441 or UNS S43940 grades. This steel is dual stabilised with titanium and niobium and is fully ferritic at all temperatures. Low interstitial content has a beneficial effect on mechanical and corrosion properties. *Ultrahigh-purity* types (C + N < 150 ppm) are the most preferred ones for the higher chromium content restricts the space of carbon and nitrogen in the ferrite matrix. Higher concentrations of these interstitials, as in intermediate-purity (150 < C + N < 800 ppm) types, initiate some difficulties when welding. Loss of toughness and ductility is expected and sheet thickness is kept low. [13]

Dual stabilisation effectively prevents sensitisation reaction but overuse of these elements may lead to secondary phase embrittlement, like Laves phases e.g. Fe<sub>2</sub>Nb [41]. Fully ferritic structure is susceptible to the HAZ grain coarsening but the dual stabilisation, mainly niobium, effectively restricts the grain coarsening by pinning the grain boundaries [34]. Unnecessary PWHTs or unsuitable operating temperatures lead to secondary phase embrittlement phenomena.





Figure 15. HAZ of grade 1.4509. Notice the limited grain coarsening in HTHAZ. [34]

Austenitic filler metals are mostly used. If using ferritic filler metals, the compositional compatibility has to be considered i.e. low interstitial content and modest content of stabilising elements. Titanium content increases the number of equiaxed grains in the weld centre, which is advantageous when comparing to typical columnar grain structure, which is vulnerable to weld cracking [13].

## 5.5 Grade 1.4521

This medium chromium steel is similar to grade 1.4509 but with extra molybdenum content to enhance the corrosion resistance. Similar matters must be emphasised, i.e. interstitial content, grain coarsening and secondary phase embrittlement phenomena. In addition, molybdenum content increases the risk to Laves phase embrittlement at temperatures 600...825 °C [41].

Molybdenum containing austenitic filler metals are mostly used. Ultrahigh-purity (C + N + O) ferritic filler metal is an option, but the availability is indefinite.

#### 5.6 Higher chromium grades

Even though high chromium grades provide the highest corrosion resistance, these usually called *superferritics* are increasingly susceptible to grain coarsening and secondary phase embrittlement phenomena. Lower interstitial contents i.e. ultrahighpurity grades are mandatory as the toughness decreases substantially with the increasing chromium content. [13] Figure 16 represents this influence.





Figure 16. Influence of interstitial content on the toughness behaviour of Fe-Cr alloys with altering chromium content:  $\triangle$  = high toughness • = low toughness [16].

As the chromium content increases, the lower C+N content is required for adequate impact toughness. When welding, the combined influence of C+N content and grain coarsening in the HAZ (Figure 4 p. 11) will decrease the toughness in the welded joint considerably. Furthermore, the higher chromium contents will more likely, and more rapidly, produce secondary phases like 475 °C embrittlement or sigma phases, than the lower chromium ones. [13]

Some examples of these higher chromium grades are 1.4621 (445), 1.4762 (446) and "E-Brite26-1". Even though heat input has a limited effect to mechanical properties of these high chromium grades, the use of low heat input is proposed for two purposes. First, the grain coarsening is kept at minimum level, although not avoided. Second, the faster cooling rate suppresses the precipitation towards the grain boundaries. These intergranular precipitates have the most detrimental effect on toughness and corrosion properties. Therefore, intragranular precipitation is more tolerable. If the cooling rate is slow, certain PWHTs can improve these properties afterwards. Temperatures should be quite high (over 850 °C), as the sigma- and Laves phase embrittlement phenomena occurs around 600 °C to 825 °C, depending on composition [41]. Water quenching should be preferred after the PWHT to prevent the intergranular precipitation. [7]

Austenitic filler metals are required since the ultrahigh-purity ferritic filler metals are not available. Typically used austenitic filler metals are ER309, ER310 or ER316L.



## **6** Conclusions

Ferritic stainless steels are a wide range of chromium-based steels. Considering weldability, many steel grades have some distinctive characteristics. Grain coarsening, formation of martensite, secondary phases and precipitates as well as interstitial content affects to the properties of these steels.

- Grain coarsening in base metal and HAZ is evident in the instance of welding ferritics. This grain growth is permanent as the ferritic structure is unable to recrystallise in PWHTs. The only effective way to minimise grain coarsening is the use of a low heat input.
- Formation of martensite concerns unstabilised steel grades. Low chromium ones, like 1.4003, have the highest martensite content, reaching up to 100 % in the HAZ, although it is majorly dependent on chemical composition. Medium chromium unstabilised ferritics like 1.4016 exhibit some amount of austenite in the HAZ but near the fusion boundary the structure becomes fully ferritic, which then leads to pronounced grain coarsening.
- High martensite content in the HAZ, roughly 90...100 %, increases the impact toughness considerably in low carbon grades. Lower martensite contents are harmful because grain boundary martensite deteriorates the toughness and ductility of the HAZ. Unstabilised grade 1.4016 (430) suffers most notably from this phenomenon. A PWHT (tempering) is recommended to retain some of the mechanical and corrosion properties.
- Although stabilised steel grades effectively prevent the sensitisation phenomenon, these grades have low impact toughness overall, mostly associated with larger grain sizes and intermediate interstitial content (over 150 ppm). Unnecessary heat treatments can also lead to various secondary phases, like for the Fe<sub>2</sub>Nb Laves phase embrittlement.
- All ferritics suffer from hydrogen embrittlement and martensite content is acknowledged to increase this threat. Any hydrogen access should be eliminated when welding ferritics. Preheating and PWHT is preferred when welding predominantly martensitic steel grades.
- Austenitic filler metals are preferred. However, some applications like exhaust systems are more suitable to be welded with ferritic filler metals.
- Argon based shielding gases should be used when welding ferritics. Oxygen or carbon dioxide additions improve the arc behaviour in MAG welding. Nitrogen or hydrogen additions should not be used.



## References

- 1 Taban, E. & Deleu, E. & Dhooge, A. & Kaluc, E. *Gas metal acr welding of modified X2CrNi12 ferritic stainless steel*. Kovove Materialy, 2007.
- 2 Kyröläinen, A. & Lukkari, J. *Ruostumattomat teräkset ja niiden hitsaus*. Metalliteollisuuden keskusliitto MET, 1999.
- 3 Shanmugam, K. & Lakshminarayanan, A.K. & Balasubramanian, V. *Effect of weld metal properties on fatigue crack growth behaviour of gas tungsten arc welded AISI 409M grade ferritic stainless steel joints*. International journal of pressure vessels and piping 86, Elsevier Ltd, 2009.
- 4 Madeira, R.P. & Modenesi, P.J. *The study of 430Ti and 430LNb ferritic welding wires for application in the cold part of automotive exhaust systems.* Welding international Vol. 24/6, 2008.
- 5 International stainless steel forum. The ferritic solution. ISSF, 2007.
- 6 Taban, E. & Kaluc, E. & Dhooge, A. *Hybrid (plasma + gas tungsten arc)* weldability of modified 12% Cr ferritic stainless steel. Materials and design, Elsevier Ltd, 2009.
- 7 Lippold, J.C. *A review of the welding metallurgy and weldability of ferritic stainless steels*. Edison welding institute, 1990.
- 8 van Warmelo, M. & Nolan, D. & Norrish, J. *Mitigation of sensitisation effects in unstabilised 12%Cr ferritic stainless steel weldments.* Materials science and engineering A, Elsevier Ltd, 2007.
- 9 Zaayman, J.J.J. Improvements to the toughness of the heat affected zone in welds of 11 to 12 per cent chromium steels. Doctorate thesis supplement, University of Pretoria, 1994.
- 10 Lindroos V. & Sulonen, M. & Veistinen, M. *Uudistettu Miekk-ojan metallioppi*. Otava, 1982.
- 11 Kou, S. Welding metallurgy. Wiley, 2003.
- 12 School of materials science and engineering. In *http://www.hsctut.materials.unsw.edu.au/Crack%20Theory/Images/ductilebrittlegr aph.png*. University of New South Wales. [Referenced 21.10.2010].



- 13 Krysiak, K.F. & Grubb, J.F. & Pollard, B. & Campbell, R.D. *Selection of wrought ferritic stainless steels*. ASM Handbook Vol. 06: Welding, Brazing, and Soldering, ASM International, 1993.
- 14 Tenhovirta, M. *Impact toughness of 12%Cr ferritic stainless steel*. Master's thesis, Helsinki University of Technology, 2009.
- 15 Ohashi, N. & Ono, Y. & Kinoshita, N. *Effects of metallurgical and mechanical factors on charpy impact toughness of extra-low interstitial ferritic stainless steels.* American Society for Testing and Materials, 1980.
- 16 Van Zwieten, A.C.T.M. & Bulloch, J.H. Some considerations on the toughness properties of ferritic stainless steels - a brief overview. International Journal of Pressure Vessels & Piping Vol. 56, 1993.
- 17 Mohandas, T. & Madhusudhan Reddy, G. & Naveed, M. A comparative evaluation of gas tungsten and shielded metal arc welds of a "ferritic" stainless steel. Journal of materials processing technology, 1998.
- 18 Munir, A. *Marine diesel engine exhaust gas system, fouling and corrosion problems.* Helsinki University of Technology, Ship Laboratory, 2007.
- 19 Potgieter, J.H. & Sephton, M. & Nkosi, Z.W. *Corrosion of hot end automotive exhaust components*. Anti-Corrosion Methods and Materials 54/3, 2007.
- 20 Leinonen, J. Hitsaustekniikka handout. University of Oulu, 2007.
- 21 Outokumpu welding handbook. Outokumpu Stainless Oy, 2010.
- 22 Nelson, T.W. & Lippold, J.C. & Mills, M.J. *Nature and evolution of the fusion boundary in ferritic-austenitic dissimilar weld metals, part 1 nucleation and growth.* Welding research supplement, Welding Journal, 1999.
- 23 Huaiyue, H. *Weldability and welding procedure of ferritic stainless steel. Conference Papers.* The fourth China International modern ferritic stainless steel & modern martensitic stainless steel conference, 2011.
- 24 Lakshminarayanan, A.K. & Shanmugam, K. & Balasubramanian, V. Effect of welding processes on tensile and impact properties, hardness and microstructure of AISI 409M ferritic stainless joints fabricated by duplex stainless steel filler metal. Journal of Iron and Steel Research 16/5, 2009.
- 25 Avesta welding manual. Avesta Welding AB, 2004.



- 26 Grönlund, E. *Teräksen kaasukaarihitsaus*. Metalliteollisuuden Kustannus Oy, 1985.
- 27 Bylund, L. (Böhler Welding) via Heikkinen, H. Personal communication 1/2011.
- 28 Inui, K. & Noda, T. & Shimizu, T. & Nagata, M. *Development of the ferritic* stainless steel welding wire providing fine grain microstructure weld metal for the components of automotive exhaust system. SAE International, 2003.
- 29 Gooch, T.G. & Ginn, B.J. *Heat-affected zone toughness of SMA welded 12%Cr martensitic-ferritic steels.* Welding Journal Vol. 69, 1990.
- 30 Panton-Kent, R. *Phase balance in 9%Cr1%Mo steel welds*. TWI Report Bulletin 1, 1991.
- 31 Zheng, H. & Ye, X. & Jiang, L. & Wang, B. & Liu, Z. & Wang, G. Study of microstructure of low carbon 12% chromium stainless steel in high temperature heat-affected zone. Materials and Design, Elsevier Ltd, 2010.
- 32 Balmforth, M.C. & Lippold, J.C. *A new ferritic-martensitic stainless steel constitution diagram.* Welding research supplement, Welding Journal No. 12, 2000.
- 33 Sánchez-Cabrera, V.M. & Rubio-González, C. & Ruíz-Vela, J.I. & Ramírez-Baltazar, C. Effect of preheating temperature and filler metal type on the microstructure, fracture toughness and fatique crack growth of stainless steel welded joints. Materials science and engineering A 452-453, Elsevier Ltd, 2006.
- 34 Anttila, S. *Mechanical and corrosion properties of ferritic stainless steels welded with ferritic filler metals.* Master's Thesis, University of Oulu, 2011.
- 35 Meyer, A.M. & du Toit, M. Interstitial diffusion of carbon and nitrogen into heataffected zones of 11-12% chromium steel welds. Welding Journal, 2001.
- 36 Lo, K.H.& Shek, C.H. & Lai, J.K.L. Recent developments in stainless steels. Materials science and engineering R, Elsevier Ltd, 2009.
- 37 Anttila, S. *Ferriittisten ruostumattomien terästen haurastumisilmiöt ja niiden lämpökäsittelyt.* Internal report TRC5979\_2010, Outokumpu Stainless Oy, 2010.
- 38 Souza, J. & Abreu, H. & Nascimento, A. & de Paiva, J. & de Lima-Neto, P. & Tavares, S. *Effects of Low-Temperature Aging on AISI 444 Steel*. Journal of materials engineering and performance, ASM International, 2004.



- 39 Yamagishi, T. & Akita, M. & Nakajima, M. & Uematsu, Y. & Tokaji, K. *Effect of σ-phase embrittlement on fatigue behavior in high chromium ferritic stainless steel*.
  Procedia engineering 2, Elsevier Ltd, 2010.
- 40 Villafuerte, J.C. & Kerr, H.W. & David, S.A. *Mechanisms of equiaxed grain formation in ferritic stainless steel gas tungsten arc welds*. Materials science and engineering A 194, 1995.
- 41 Sello, M. P. & Stumpf, W.E. *Laves phase embrittlement of the ferritic stainless steel type AISI 441*. Materials Science and Engineering A, 2010.
- 42 Kivirinta, S. Ferriittisen 17 % Cr sisältävän ruostumattoman teräksen herkistymisasteen määrittäminen. Master's thesis, Tampere University of Technology, 2007.
- 43 Karjalainen, P. *Hitsausmetallurgia handout*. University of Oulu, 2005.
- Kotecki, DJ. Welding of stainless steels; welding consumables and procedures.
  ASM Handbook Vol. 06: Welding, Brazing, and Soldering, ASM International, 1993.
- 45 du Toit, M. & Naudé, J. *The influence of stabilization with titanium on the heataffected zone sensitization of 11 to 12 % chromium ferritic stainless steels under low heat input welding conditions.* Welding in the World Vol. 55, 2011.
- 46 Kim, J.K. & Kim, Y.H. & Uhm, S.H. & Lee, J.S. & Kim, K.Y. *Intergranular corrosion of Ti-stabilized 11 wt% Cr ferritic stainless steel for automotive exhaust systems*. Corrosion science 51, Elsevier Ltd, 2009.
- 47 van Niekerk, C.J. & du Toit, M. *Sensitization behaviour of 11-12% Cr AISI 409 stainless steel during low heat input welding*. The Journal of The Southern African Institute of Mining and Metallurgy, 2011.