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Welding stainless steel in Shipbuilding

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INTRODUCTION

1.1 HISTORY ^[1] ^[2] ^[3] ^[4]



Figure 1.1 An announcement in the 1915 *New York Times*, of the discovery of stainless steel.

There are only a few corrosion-resistant iron artifacts that survive from antiquity. Unlike stainless steel though, these artifacts owe their durability to their high phosphorus content and not to chromium. Finding itself in favorable local weather conditions, phosphorus would promote the formation of a solid protective passivation layer of iron oxides and phosphates, rather than the non-protective, cracked rust layer that develops on most ironwork.

The French metallurgist Pierre Berthier was the first to recognize the corrosion resistance of iron-chromium alloys in 1821. He noted their resistance against attack by some acids and suggested they could be used in making cutlery. The combination of low carbon and high chromium found in modern stainless steel, however, was too difficult to be produced by metallurgists back in the 19th century. The high-chromium alloys they could produce were too brittle to be of any practical use.

At the dawn of 1890, it was Hans Goldschmidt of Germany who developed an aluminothermic (thermite) process for producing carbon-free chromium. During 1904 and 1911, Leon Guillet of France, as well as several other researchers, managed to prepare alloys that would today be considered stainless steel. Another development that enabled stainless steel widespread production was the fact that in 1899 Paul Heroult of France managed to develop the direct-arc electric melting furnace.

Many more developments took place all over the world in the years that followed, with Christian Dantzen and Frederick Becket industrializing ferritic stainless steel in the United States, but it wasn't until October 17, 1912, when Krupp engineers Benno Strauss and Eduard Maurer actually patented austenitic stainless steel.

However, the inventor of stainless steel is commonly considered to be Harry Brearley of the Brown-Firth research laboratory in Sheffield, England. While he was seeking an erosion-resistant alloy for gun barrels in 1913, he accidentally discovered and subsequently industrialized a martensitic stainless steel alloy of composition 12.86wt% Cr, 0.24wt% C, 0.20wt% Si and 0.44wt% Mn. The discovery was announced in a January 1915 newspaper article in The New York Times and was later marketed under the "Staybrite" brand by Firth Vickers in England, thus marking the start of the age of stainless steel.

1.2 DEFINITION ^{[1] [2] [3] [4]}

All metals except gold, platinum and palladium corrode spontaneously, and while stainless steel is a name that correctly signifies the metal's rust resistant properties, this metal is not 100% stain-proof and therefore it is not completely immune to corrosion and rust. To better understand this concept we should first understand the source of this metal's resistance.

Stainless steel may be defined as a family of high-alloy steels containing a minimum of 10.5wt% chromium, although the normal chromium content of these steels ranges from 10.5wt% to 11.5wt%. This is the amount of chromium required to allow, upon contact with oxygen, the formation of a natural barrier of chromium oxide, called a "passive film". This passive film has the ability to self-heal whenever it is breached, creating a new film as the exposed chromium interacts with oxygen.

However, since the metal's corrosion resistance is derived from its passive film, abrasive materials and corrosive media that attack and remove the chromium oxide can cause corrosion of stainless steels. Keeping an item made of stainless steel clean is the best way to prevent it from being corroded, but one should refrain from using brushes or chloride-containing detergents when cleaning such items.

It should also be pointed out that many Cr steels are also liable to exhibit "rust", if some of the Cr is tied up as carbides. This reduces the Cr composition of the matrix to a degree that does not support the formation of the aforementioned passive film.

Another function that chromium content besets upon stainless steel is the metal's ability to resist oxidation even at high temperatures. Some alloys can be utilized at temperatures as high as 1000°C (1850°F), something that earned them the name of "heat-resisting alloys". This also led to the development of stainless steel alloys of medium Cr content (approx. 16wt%) and high Ni content (about 35wt%) that have the ability to resist carburization.

Because of its resistance to corrosion and staining, low maintenance, relative inexpense, familiar luster and 100% recyclability, stainless steel has become an ideal base material for a host of commercial applications. Stainless steel can be found in everything; from buildings and wrist watches to monuments and skate blades. Most people encounter stainless steel products every day. The corrosion resistance and antibacterial properties of the alloy make it the perfect material for food transport, storage tankers and food processing plants. Stainless steel is milled into coils, sheets, plates, bars, wire, and tubing and subsequently is used in cookware, cutlery, hardware, surgical instruments, major appliances, industrial equipment, and even as an automotive, aerospace and marine structural alloy.

1.3 TYPES OF STAINLESS STEEL ^{[1] [2] [3] [4]}

The ample variety of stainless steel makes it the perfect alloy to use in a wide range of applications requiring adverse properties. When nickel is added, for instance, the austenite structure is stabilized. If hardness and strength is required, carbon should be added. Large quantities of manganese are also being used in some compositions of stainless steel, since manganese preserves an austenitic structure in the steel, just as nickel does, but at a lower cost.

Unlike other metals, stainless steels are not categorized by their composition. Their classification is based on their predominant crystalline structure. As such, there are four main types of stainless steel:

- **Austenitic (2XX, 3XX series):** these are the most commonly used stainless steels, since they comprise more than 70% of the entire stainless steel production. Superaustenitic stainless steels (such as alloy AL-6XN and 254SMO) are known for their resistance to chloride pitting and crevice corrosion. These properties are a derivation of the alloy's composition of high molybdenum (>6%), and nitrogen additions. Higher nickel content can also ensure higher resistance to stress-corrosion cracking, than that of the 3XX series. However superaustenitic stainless steels tend to be much more expensive than their 3XX-series counterparts, because of their high alloy content. Austenitic types are essentially nonmagnetic.
- **Ferritic (4XX series):** the ferritic types are highly corrosion-resistant, but less durable than austenitic grades. Ferritic stainless steels are ferromagnetic, since a small amount of residual ferrite is enough to introduce a slight ferromagnetic condition, although not as strong as in a magnetic material.
- **Martensitic (4XX series):** these steels are less corrosion-resistant than austenitic and ferritic, but they are much stronger and tougher than them. They also tend to be highly machineable and can be further hardened by heat treatment, although they are a bit more brittle due to their high carbon content. Martensitic types, just like their ferritic cousins, are ferromagnetic too.
- **Duplex:** duplex stainless steels have a mixed microstructure of austenite and ferrite. Usually the mix is 40%/60% respectively, but the common aim is to produce a mix in the area of 50%/50%. These types show greater strength than austenitic stainless steels and they have improved resistance to certain types of corrosion, namely localized types of corrosion. The most commonly used duplex stainless steels are 2205 and 2507, the latter being also called "SuperDuplex" due to its higher corrosion resistance.

- **Precipitation hardened (PH):** these types of stainless steel have corrosion resistance that is highly comparable to the austenitic varieties, with the added benefit of achieving even higher strengths than the other martensitic grades through precipitation hardening. There is a trend in many projects to opt for an ultra-high-strength stainless steel when possible, since a large percentage of their budget is spent on dealing with corrosion.

The different types of stainless steel show many variations in their physical and mechanical properties, and this is something that reflects greatly on their welding characteristics.

1.4 MANUFACTURING METHODS ^{[2] [5]}

In order to produce stainless steel an electric arc furnace is used to melt together all the raw materials. The melting process usually requires 8 to 12 hours of intense heat. The molten charge is then transferred to a refining vessel, where the process of argon-oxygen decarburization (AOD) or vacuum-oxygen decarburization will take place. The use of one of these processes is deemed necessary since the molten charge normally contains from 1.5 to 2wt% carbon. Injecting a mix of argon and oxygen into the molten charge, forces the carbon to combine with the oxygen, forming carbon monoxide which is then liberated from the molten metal. Argon is used in to control the argon and oxygen mixture and achieve the desired carbon content.

The VOD process shows many similarities, except from the fact that the argon carrier gas is not necessary anymore and oxygen is injected directly into the molten metal. As an added benefit, these processes also drastically reduce the residual sulfur content to as low as 0.001wt%.

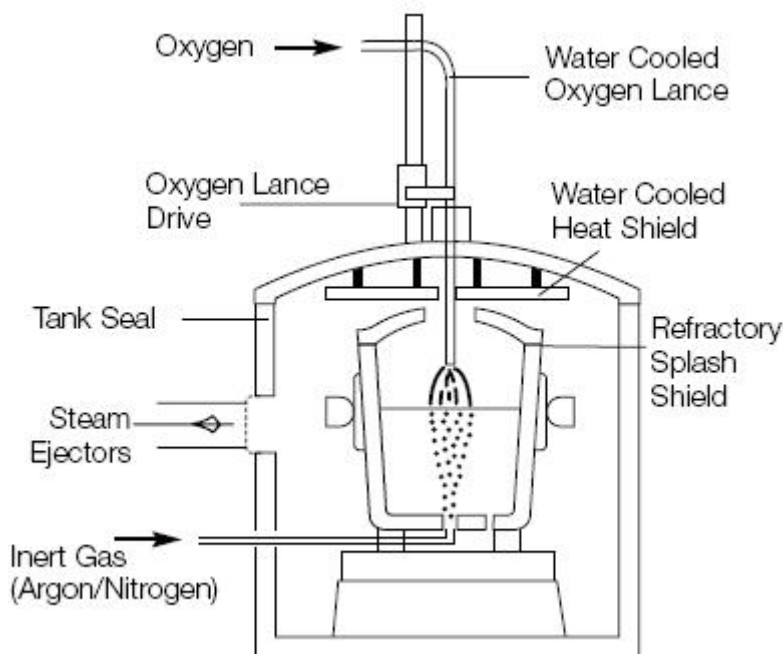


Figure 1.2 A vacuum-oxygen decarburization furnace.

At this point stainless steel can be cast into semi-finished forms, which go through further forming operations.

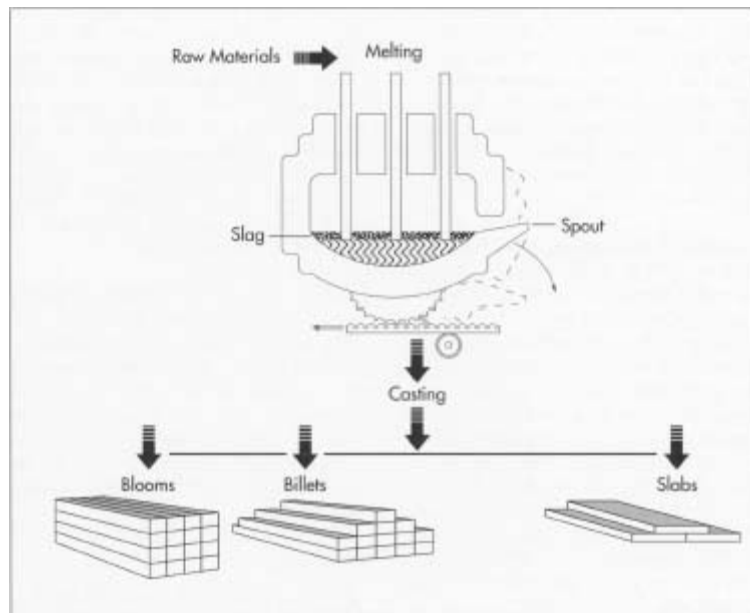


Figure 1.3 The VOD method for manufacturing stainless steels.

Beginning with hot rolling, in which stainless steel is heated and passed through huge rolls, blooms and billets are formed into bars and wire, while slabs are formed into plates, strips and sheets. After the stainless steel is formed, most types have to go through an annealing process, in which the metal is heated and cooled under controlled conditions. This treatment relieves internal stresses and softens the metal. The aforementioned heat treatment is also known as **age hardening** and requires extremely careful control, as any miscalculation (not so much in the heating rate to reach the aging temperature, as much as in the cooling rate) can seriously affect the metal's properties. Lower aging temperatures produce high strength alloys with low fracture toughness. Higher aging temperatures produce a lower strength, but tougher material. A post-aging quenching treatment (rapid cooling) can increase the alloy's toughness without significant loss in strength. The type of heat treatment depends on the type of steel: austenitic, ferritic or martensitic stainless steel. It should be also noted that if the alloy is cooled too slowly, carbide precipitation can occur, but this build-up can be eliminated by thermal stabilization (holding the metal for several hours at 815 to 871°C (1500 to 1600°F).

Annealing sometimes causes a scale or build-up to form on the metal and this has to be removed before going through any of the next steps of the manufacturing. Pickling is one of the most common methods and it uses a nitric-hydrofluoric acid bath to descale the steel. Another method is electrocleaning, when an electric current is applied to the surface using a cathode and phosphoric acid and the scale is removed.

After the annealing and descaling processes different cutting operations are usually necessary to obtain the desired blank shape or size. Mechanical cutting is accomplished by a variety of methods; straight shearing using guillotine knives, circle shearing using circular knives, sawing using high-speed steel blades, blanking using punches and dies to punch out the shape by shearing and nibbling, which is a process of cutting by blanking out a series of overlapping holes. Flame cutting and plasma jet cutting can also be used on stainless steel and is considered a rather clean and fast method.

ALLOYING ELEMENTS

2.1 GENERAL INFORMATION ^{[2] [16] [17] [18]}

Despite its abundance and extended versatility, iron may be too soft to be used for the purpose of structure and is easily susceptible to rust. Because of this, rusted iron loses integrity and becomes highly unreliable in critical situations.

However, the addition of small quantities of other elements greatly enhances its mechanical properties. This is because the different atom sizes of other elements interrupt the orderly arrangement of atoms in the lattice and prevents them from easily sliding over each other.

These 'other elements' are called **alloying elements** and have the capability to block slip planes. The principal alloying additions to stainless steel are chromium and carbon for the ferritic and martensitic grades, with the addition of nickel for the austenitic and duplex grades. Practically all stainless steels contain manganese and silicon as additions, since these alloying elements have beneficial effects on stainless steels. Other alloying additions include molybdenum, niobium, titanium, aluminum, copper, tungsten, nitrogen and other elements. All of these elements may have a beneficial effect on fabricability, corrosion resistance, or to influence microstructure. Impurity elements commonly found in stainless steels include nitrogen, oxygen, sulphur and phosphorus. All of these alloy and impurity elements have some effect on weldability and performance.

The following section provides some insight into the influence that some of these alloying elements have on the properties and performance of stainless steels.

2.2 CARBON (C) ^{[2] [16] [17] [18]}



Carbon is a non-metallic interstitial alloying element, important in all ferrous metal-based materials. Carbon is always present in metallic alloys, i.e. in all grades of stainless steel and heat resistant alloys. It is a very strong austenitizer and increases the strength of steel.

In austenitic, ferritic, and duplex stainless steels, it is kept to low levels (typically 0.005% C to 0.03% C in low carbon grades) to retain the desired properties and mechanical characteristics.

In martensitic stainless steels, carbon is deliberately added to obtain both high strength and high hardness, thanks to the formation of a martensitic structure. However, with increased strength comes decreased ductility, thus a greater tendency for fracture. It is not possible to anneal a high carbon stainless steel to a low hardness level, without using “L” grade alloys. In martensitic grades, carbon is added as an alloying element, in amounts varying from 0.15wt% to 1.2wt% to make these alloys heat treatable by quenching.

The principal effect of carbon on corrosion resistance is determined by the way in which it exists in the alloy. Carbon can diffuse rapidly through the structure and concentrate on the grain boundaries. Therefore, it can be combined with chromium as a separate constituent (chromium carbide) and have a detrimental effect on corrosion resistance by removing some of the chromium from solid solution in the alloy and, as a consequence, reducing the amount of chromium available to ensure corrosion resistance. This adverse effect of carbon can be caused if the alloy is cooled too slowly after hot working or annealing or subsequently reheated (as in welding operations). The consequence is an unwanted precipitation of carbide-containing chromium. This precipitation of carbon takes place at grain boundaries and is referred to as sensitization. It has been demonstrated that the loss of chromium associated with the carbide precipitation lowers corrosion resistance and brings about greater susceptibility to localized corrosion, i.e. intergranular corrosion, at the grain boundaries following the network of chromium carbides.

2.3 CHROMIUM (Cr) ^{[2] [16] [17] [18]}



About 85% of the chromite mined is used in metallurgy, namely stainless steels, low-alloy steels, high-strength alloy steels, tool steels and high-performance alloys such as chromium-cobalt-tungsten (or molybdenum) alloys, nickel-chromium-manganese-niobium-tantalum (or titanium) alloys, nickel-chromium-molybdenum alloys, cobalt-chromium alloys and some maraging steels (high-strength alloy irons of the precipitation hardening type). Due to its strength and its high resistance to corrosion, chromium is often used in plating and metal finishing.

The properties that distinguish stainless steels i.e. Fe-Cr-(Mo) alloys and Fe-Cr-Ni-(Mo) alloys from other corrosion-resistant materials depend essentially on chromium. The high degree of reactivity of chromium is the basis for the effectiveness of chromium as an alloying element in stainless steels. The resistance of these metallic alloys to the chemical effects of corrosive agents is determined by their ability to protect themselves through the formation of an adherent, insoluble and self-repairing film of reaction products (chromium oxide) that shields the metal substrate from uniform and localized attack. This oxide has a stoichiometry of $(\text{Fe,Cr})_2\text{O}_3$ and is called passive layer or passive film; it is a very fine layer on the surface, of the order of 1.0 to 2.0 nm, which reduces the corrosion rate to negligible levels and has a structure similar to chromite. For passivation to occur and remain stable, the Fe-Cr alloy must have a minimum chromium content of about 10.5wt%, above which steel is considered “stainless” under ambient conditions. The corrosion resistance of Fe-Cr alloys tends to improve as the chromium content is increased, and, in fact, higher levels of chromium may be required for oxide stability in more aggressive environments.

Chromium is a strong carbide former. The most common Cr-rich carbide is M_{23}C_6 and it is found in virtually all the stainless steels. It is also possible for a Cr_7C_3 carbide to be formed, although this carbide type is not common. Other complex carbides and carbonitrides $[\text{M}_{23}(\text{C,N})_6]$ are also possible.

Chromium is also a key ingredient in the formation of intermetallic compounds, many of which tend to cause embrittlement to stainless steels. The most common is σ -phase, which in the Fe-Cr system is a (Fe,Cr) compound that forms below 815° C. Σ -phase can form in virtually any stainless steel, but tends to be most common in austenitic, ferritic, and duplex alloys with high chromium content. Chromium is also present in chi (1) and Laves intermetallic phases.

2.4 NICKEL (Ni) ^{[2] [16] [17] [18]}



Stainless steels use almost 65% of the entire nickel production, compared to 45% ten years ago. High-performance alloys (nickel-based, cobalt-based, and iron-nickel-based high-performance alloys) represent another growing metallurgical end-use for nickel.

The primary function of nickel is to promote the austenite phase so that predominantly austenitic or austenitic-ferritic alloys can be produced. By adding sufficient nickel, the austenite phase field can be expanded to the point that austenite is stable at room temperature and below.

Nickel has no direct influence on the passive layer that protects stainless steels. It does exert, however, a beneficial effect, particularly in sulphuric acid environments. Stainless steels owe their wide range of mechanical properties to nickel; a set of mechanical properties that make stainless steel alloys unparalleled by any other alloy system today. A perfect example are the austenitic stainless steels, typified by 18-8, where the tendency of Nickel to form Austenite is responsible for a great toughness (impact strength) and high strength at both high and low temperatures. What is remarkable is that these properties are retained to as low as cryogenic temperatures.

Nickel also greatly improves resistance to oxidation and corrosion. Nickel is added in large amounts, over about 8wt%, to high Chromium stainless steels to promote resistance to corrosion of the nickel-based alloys, even under conditions where the passive layers may be absent, or may be destroyed locally or uniformly. For instance, pitting corrosion tends to progress less rapidly in high-nickel alloys.

Nickel forms the base of high temperature super-alloys because of its ability to develop an adherent oxide and precipitation hardening phases based on Ni_3Al . Nickel is a moderate strengthener, and consequently large amounts can be added to low-alloy steel before strength increases to an undesirable level. In low-alloy steel, nickel appears to have a greater overall, beneficial effect on toughness transition temperature than any other substitutional alloying element.

On the off-hand, it should also be noted that nickel has been associated with a decrease in stress corrosion cracking (SCC) resistance.

2.5 MANGANESE (Mn) ^{[2] [16] [17] [18]}



Another element that aids weld penetration is manganese. The ideal range is between 1.5wt% and 1.75wt%. Manganese has three key properties that make it essential to steel making, by aiding the stability of stainless steel: its ability to increase the solubility of nitrogen, the fact that it can combine with sulphur and its powerful deoxidation capacity. When added to molten steel, manganese will react with oxygen to form manganese oxide (MnO).

The only negative aspect of manganese is its tendency to form inclusions in the steel, especially in the presence of sulphur. For this reason some semiconductor applications limit manganese to 0.5wt% maximum or in some cases to 0.05wt% maximum.

Besides entering into combination with oxygen and sulphur, manganese can also influence the behaviour of steel. The presence of manganese increases the strength, toughness and hardenability of the steel and it improves its hot working properties. It should be also noted that manganese is an austenite promoter as well and very large additions of manganese, about 12wt% to 15wt%, make the steel austenitic even at room temperature.

2.6 SILICON (Si) ^{[2] [16] [17] [18]}



Silicon is a metalloid (non-metallic element) produced by the reduction of silica (SiO_2) at high temperatures.

Silicon is added to the steel to assist deoxidation following AOD refining and to improve fluidity of the molten metal; it is a ferrite stabilizer. In small amounts, silicon confers mild hardenability on steels. Small amounts of silicon and copper are usually added to the austenitic stainless steels containing molybdenum to improve corrosion resistance in sulphuric acid.

Silicon has several problems associated with its presence in steels, such as “sagging” of the weld and formation of weld slag spots, especially if the cover gas is contaminated with oxygen or if moisture is present. It is also known to form a number of iron silicides (FeSi , Fe_2Si , Fe_3Si , Fe_5Si_3) and a Cr_3Si intermetallic, all of which tend to embrittle the structure. It also expands the composition range over which σ -phase forms.

Most of the common austenitic stainless steels limit silicon at 0.75wt%. If the silicon becomes too high it will promote ferrite formation, and during cold working form transformation martensite. However, high-silicon content in austenitic stainless steels not only improves resistance to oxidation but also prevents carburizing at elevated temperatures. The presently accepted silicon range is 0.4wt% to 0.5wt%.

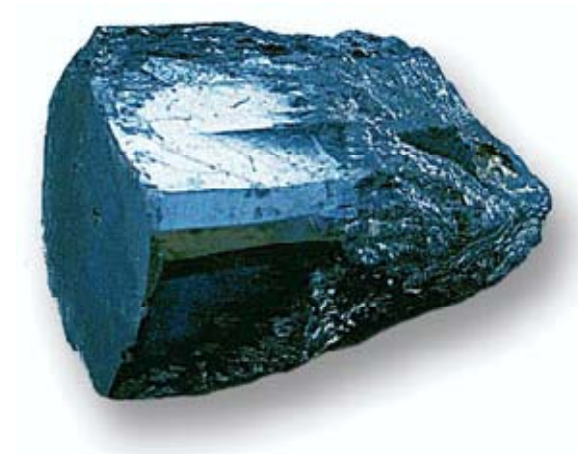
2.7 MOLYBDENUM (Mo) ^{[2] [16] [17] [18] [19]}



The high melting point of molybdenum makes it important for giving strength to steel and other metallic alloys at high temperatures. However, this can also have a negative effect, since alloys containing molybdenum will be more difficult to hot work. Molybdenum is a carbide former and the addition of as little as 0.5wt% increases the secondary hardening characteristics of the steel, thus resulting in higher room-temperature yield and tensile strength and improved elevated-temperature properties. Molybdenum is a ferrite promoter, and its presence will promote ferrite formation and retention in the microstructure. This can potentially be a problem in martensitic grades, where residual room-temperature ferrite can reduce toughness and ductility.

Molybdenum is also added to metallic alloys because of its resistance to corrosion. Molybdenum is used in ferritic, austenitic and duplex stainless steels in amounts up to 8wt% or more in super austenites. Most commonly it is used in the range from 2wt% to 4wt%. Even such relatively small percentages of molybdenum have powerful effects in improving the resistance to pitting corrosion in chloride environments and to crevice corrosion in both Fe-Cr alloys and Fe-Cr-Ni alloys. Molybdenum reduces the intensity of the oxidizing effect required to insure passivity and decrease the tendency of previously formed passive films to break down.

2.8 NIOBIUM (Nb) ^{[2] [16] [17] [18]}



Niobium is added to steel in order to stabilize carbon, and, as such, performs in the same way as described for titanium. The most important application for niobium is an alloying element for micro-alloyed steels in which it is used to strengthen the metal structure. The strengthening mechanisms are based on the precipitation of fine dispersed carbide (NbC).

An important secondary application for niobium is in super alloys operating in the hot section of aircraft engines. Niobium is also utilized in stainless steels (austenitic and ferritic) as a stabilizing element and to improve creep resistance (mainly in ferritic steels) and for purposes involving superconductivity (niobium – titanium alloys), which permitted the development of modern magnets.

In stainless steels, as far as corrosion resistance is concerned, it is well known that stabilizing the grade by Nb additions prevents the risk of intergranular corrosion in heat affected zones. To prevent this, the niobium is added in sufficient amounts, depending on the carbon and nitrogen (ferritic types) levels. In ferritic stainless steels, the addition of niobium is one of the most effective methods for improving thermal fatigue resistance.

2.9 TITANIUM (Ti) AND ZIRCONIUM (Zr) ^{[2] [16] [17] [18]}



Titanium and/or zirconium are popular elements for deoxidizing the melt after AOD refining. Deoxidation is necessary to prevent fracture of the ingot during hot rolling, so the steel mills want totally “killed” steels. Titanium is added for carbide stabilization especially when the material is to be welded. Because oxygen promotes weld penetration, excess titanium and/or zirconium will reverse convection currents and cause variations in weld uniformity. Longitudinal welds will develop arrowheads, a slag spot will form, and then the weld becomes normal again. In addition, excess titanium and/or zirconium scavenge oxygen from the cover gas and forms slag spots on the weld.

Titanium combines with carbon and nitrogen to form titanium carbides (TiC) and titanium nitrides (TiN) respectively. Titanium carbides are quite stable and hard to dissolve in steel, which tends to minimize the occurrence of inter-granular corrosion. In the presence of sulphur, titanium reacts to form stable sulphides, Ti_2S , which have been shown to form in preference to manganese sulphides (MnS) which are known to act as pit initiation sites. As a consequence, pitting corrosion resistance is improved.

It appears that titanium/zirconium levels in the range of 0.001wt% to 0.004wt% have little effect on the weld quality, but when the concentration is in excess of 0.007wt% carbon combines with titanium in preference to chromium, preventing a tie-up of corrosion-resisting chromium as inter-granular carbides and the accompanying loss of corrosion resistance at the grain boundaries.

In low alloy steels, titanium has a strong desire to unite with carbon, nitrogen and oxygen. When dissolved in steel, titanium is believed to increase hardenability; however, the carbide-forming tendency of this element is so strong that it is frequently in the steel structure as undissolved carbides and in this way decreases hardenability.

2.10 ALUMINUM (Al) ^{[2] [16] [17] [18]}

Aluminum is another element that is used to deoxidize the steel. It acts much the same as titanium both as a slag former and its tendency to form arrowheads on the weld with an associated slag spot. Aluminum in the range of 0.006wt% appears to be safe, but residuals in excess of 0.010wt% may cause slag formation.

2.11 COPPER (Cu) ^{[2] [16] [17] [18]}

Copper is normally present in stainless steels as a residual element. In the range of 3wt% to 4wt% copper improves corrosion resistance to sulphuric acid and sea water environments. At low levels and in combination with other elements copper will produce age hardening. But the copper content is creeping upward because it is being introduced through the use of scrap metal during compounding the melt, and AOD refining does not eliminate it. Today, the copper content of most of the common stainless grades, 304, 304L, 316, 316L, is in the range of 0.4wt% to 0.5wt%. Ten years ago they were half that. Copper may not be all-bad, since it improves resistance to sulphuric acid corrosion. However, if the alloy is to be used for any type of human implant it may cause problems.

2.12 NITROGEN (N) ^{[2] [16] [17] [18]}

Nitrogen, as carbon, is an interstitial element in steel because its atomic size is sufficiently small in relation to that of iron, chromium or nickel. This allows the element to enter both the α - and γ -alloy lattices as interstitial solute atoms. The solubility of nitrogen is greater in austenite than in ferrite, because of the larger interstices available. At the temperature of 1100° C, nitrogen solubility in austenite is as high as 2.4wt% while in ferrite is only 0.1wt%.

Nitrogen has the effect of increasing the austenite stability of stainless steels and is, as in the case of nickel, an austenite forming element. Yield strength is greatly improved when nitrogen is added to stainless steels. When added in austenitic and duplex stainless steels, nitrogen increases their resistance to localized corrosion like pitting and intergranular. This is due to the precipitation of Cr_2N nitride instead of Cr_{23}C_6 carbide.

Low-carbon steel grades with up to 0.2wt% nitrogen addition were developed as a solution to the problem of low-carbon grades having lower yield strength than that of standard grades. Nitrogen in solid solution raises the yield strength to at least the same level as in the standard austenitic grades.

2.13 COBALT (Co) ^{[2] [16] [17] [18]}



Cobalt is alloyed with iron, nickel and other metals to make an alloy with special magnetic properties. Cobalt is a major constituent of the “Super Alloys”.

The uniform corrosion (<0.05 mm/yr) and pitting corrosion resistances of cobalt containing alloys tend to be excellent at about 100° C in 10% solutions of acetic acid,

ammonium chloride, ammonium sulphate, ferric chloride, formic acid, nitric acid, phosphoric acid or sodium chloride.

The cobalt matrix has a high-work hardening sensitivity, which combines with the carbide fraction and allows it to achieve excellent wear resistance associated with a high degree of corrosion resistance. Cobalt is a valuable alloying element for high-speed tool steels. It has the effect of raising ferrite to softening temperature so that tools made from cobalt-alloy steel can operate at high temperatures, maintaining their cutting capacity.

Despite all its beneficial effects however, cobalt becomes highly radioactive when exposed to the intense radiation of nuclear reactors, and, as a result, any stainless steel that is in nuclear service will have a cobalt restriction, usually at a maximum of 0.2wt%. This problem is emphasized because there is normally residual cobalt content in the nickel used in producing austenitic stainless steels.

2.14 CALCIUM (Ca) ^{[2] [16] [17] [18]}

Calcium is a strong deoxidizer and is used by some steel mills to kill the melt. Most of the calcium comes from the slag, refractories and mold powders during melting and casting. It is associated with boron in some grain boundary precipitates. These compounds form very slaggy welds. Most steels run in the range of 0.0010wt% to 0.0020wt%, but occasionally a very slaggy weld will have 0.0145wt% to 0.00175wt% calcium. The lower limits do not appear to be a problem, but the higher levels are. It is also customary to add calcium in small amounts to improve the alloy's machinability, without the detrimental effects on other properties caused by sulphur, phosphorus and selenium.

2.15 BORON (Bo) ^{[2] [16] [17] [18]}

Boron is one of those elements that some steel mills love and other mills avoid. Boron is added to the melt for sliver control during hot rolling. Boron is not an element that builds up as a residual. It is easy to tell which mills use boron because their heats are in the range of 0.0070wt% to 0.0185wt%. The solubility limit is reported to be in the range of 0.010wt%. Steel from the mills that avoid the use of boron is in the range of >0.0001wt% to 0.0006wt%. Boron is impossible to identify using microprobe analysis because of its low atomic number. The only reliable analytical method is inductively coupled plasma spectroscopy with the appropriate standards. Standard emission spectroscopy does not appear to detect boron with the same sensitivity as ICP spectroscopy.

2.16 SULPHUR (S) ^{[2] [16] [17] [18]}

Weld penetration difficulty can occur when the sulphur content is too low. During AOD (Argon Oxygen Decarburization) refining the residual sulphur is reduced to 0.001wt% or less. This is desirable to produce low inclusion steel, but very bad for weld penetration. In recent years the BPE (Bio-Pharmaceutical Equipment) Committee of the ASME has specified the sulphur content of all steel components for the pharmaceutical industry to have a sulphur content of 0.005wt% to 0.017wt%. This means the steel must be resulphurized at the time of casting. Most of these steels have a mean sulphur range of 0.012wt% to 0.014wt%. No adverse problems have been identified with sulphur at these low concentrations except a tendency to

have a higher inclusion count. Inclusions are insoluble non-metallic “dirt” in the steel. One school of thought says these inclusions may lead to a higher pitting corrosion rate.

Sulphur, together with controlled Manganese and Silicon, control the convection currents in the weld pool so the heat is carried deep into the weld. Elements that reverse the convection currents producing low penetration welds include calcium, titanium, aluminum, magnesium and the rare earths.

When added in small amounts sulphur can improve machinability, just like calcium does.

2.17 MAGNESIUM (Mg) ^{[2] [16] [17] [18]}

Like titanium and aluminum, magnesium is a strong deoxidizer. In the past it was used in some deoxidizer compounds, but today it is normally not added to the melt for deoxidization. It shows up in the steel from the slag and refractory lining of the AOD converter, ladle and caster. All steel mills have about the same amount of residual magnesium, in the order of 0.006wt% to 0.008wt%. This does not appear to be a problem.

2.18 RARE EARTHS ^{[2] [16] [17] [18]}

The common rare earths include cesium, yttrium and lanthanum. They are powerful deoxidizers and result in lack of weld adherence of the joint surfaces and weld cracks. Those alloys that have intentional additions are variable in weldability, exhibit weld cross cracks and may show balling of the weld metal on the surface. The levels where problems develop are in the range of 0.005wt% of any of the three elements. Thankfully, very few mills use rare earth additions. None of them should.

2.19 PHOSPHORUS (P) ^{[2] [16] [17] [18]}

Phosphorus is usually added with sulphur, to improve machinability. The phosphorus present in austenitic stainless steels increases strength; however, it has a detrimental effect on corrosion resistance and increases the tendency of the material to crack during welding.

2.20 SELENIUM (Se) ^{[2] [16] [17] [18]}

Selenium was previously used as an addition to improve machinability.

2.21 TUNGSTEN (W) ^{[2] [16] [17] [18] [19]}

The addition of tungsten in stainless steels affects the nature of the protective passive film. The presence of tungsten in stainless steels produces a protective film rich in chromium and tungsten oxides, which increases the alloy's overall resistance. The addition of tungsten in stainless steel leads to the formation of WO_3 in the superficial zone of the passive layer, in addition to other constituents observed in the alloy. This compound is insoluble in acidic environments and reinforces the stability of Cr_2O_3 .

2.22 FERRITE- VERSUS AUSTENITE-PROMOTING ELEMENTS AND THEIR EFFECT ON STAINLESS STEEL PROPERTIES ^{[2] [16] [17] [18]}

Alloying elements added to stainless steel can be divided into those that promote, or stabilize, either the ferrite or the austenite phase. It should be clarified that martensite is not considered a different phase (like ferrite or austenite); instead it is a transformation product that forms from austenite when cooled from elevated temperature. That means that if no austenite is formed at elevated temperatures, no martensite can form at low temperatures.

It is easily understood that austenitic stainless steels contain austenite-promoting elements that make the formation of austenite a priority, thus stabilizing the austenite phase at room temperature and below. Ferritic stainless steels, on the other hand, contain ferrite-promoting elements that make ferrite the predominant metallurgical phase present. Some stainless steels that are austenitic at elevated temperatures are fairly unstable and transform into martensite upon cooling to room temperature.

The bottom line is that by balancing austenite- and ferrite-promoting elements in stainless steels we have great power over controlling their microstructure. It should be noted, however, that such a balance has important implications in respect to mechanical properties, corrosion resistance and weldability.

The following table groups some of the alloying elements into austenite- and ferrite-promoters:

Austenite-promoting elements:	Nickel
	Manganese
	Carbon
	Nitrogen
	Copper
	Cobalt
Ferrite-promoting elements:	Chromium
	Molybdenum
	Silicon
	Niobium
	Titanium
	Aluminum
	Vanadium
	Tungsten

A general guideline as to the effect of various alloying elements on the properties of stainless steels shows on the following table, with “V” marking a beneficial effect and “X” marking a detrimental effect:

Property	C	CR	Ni	S	Mn	Si	P	Cu	Mo	Se	Ti or Nb
Corrosion Resistance	-	✓	✓	X	-	-	✓	-	✓	-	-
Mechanical Properties	✓	✓	-	-	✓	✓	✓	✓	✓	-	✓
High Temperature Resistance	-	✓	✓	X	-	-	-	-	✓	-	✓
Machinability	X	X	-	✓	-	-	✓	-	-	✓	-
Weldability	X	X	-	X	✓	-	X	-	✓	-	✓
Cold Workability	X	X	✓	X	-	-	-	✓	-	-	-

Figure 2.1 The effect of various alloying elements on the properties of stainless steels.

3

PHASE DIAGRAMS

3.1 GENERAL INFORMATION ^{[6] [7]}

A phase diagram is a type of chart used to describe conditions at which thermodynamically-distinct phases can occur at equilibrium. The **lines of equilibrium** or **phase boundaries** are part of a phase diagram's common components. These lines mark conditions under which more than one phases can coexist at equilibrium. Phase transitions will occur along those lines.

On many diagrams one could encounter points where lines of equilibrium intersect. These are called **triple points** or **eutectic (eutectoid) points** and mark the conditions at which three different phases can coexist. The difference between a eutectic and a eutectoid point is that in the former case there is a distinct liquid phase, while in the latter case the liquid phase has been replaced by a (second) solid phase. In every other aspect these two definitions are identical.

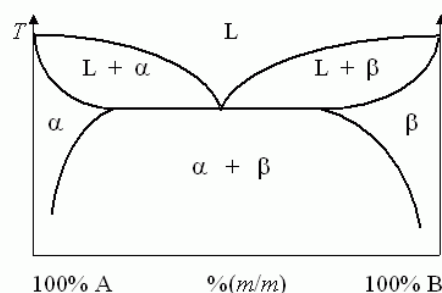


Figure 3.1 A phase diagram for a binary system displaying a eutectic point.

Another point of interest common in many (if not all) phase diagrams is the **critical point**. The critical point marks the conditions beyond which the distinction between the various phases becomes problematic.

Some other main components on many phase diagrams are the **solidus** and **liquidus lines**; solidus is the temperature below which the substance is stable in a solid state, whereas liquidus is the temperature above which the substance is stable in a liquid state.

There are many types of phase diagrams; a **2D phase diagram** is the simplest phase diagram of a single simple substance, with its two axes normally corresponding to two thermodynamic values, such as pressure and temperature. A **3D phase diagram** is nothing more than a three-dimensional representation of three thermodynamic quantities. In such a diagram the equilibrium conditions would be shown as a 3D curved surface with areas for solid, liquid, vapor phases and areas where solid and liquid, solid and vapor, or liquid and vapor can coexist in equilibrium. A line on this surface, called a **triple line**, is where solid, liquid and vapor can coexist in equilibrium. The critical point remains a point on the surface even on a 3D phase diagram. An orthographic projection of the 3D graph, showing only two of the thermodynamic quantities as the vertical and horizontal axes, effectively collapses the 3D plot into a 2D phase diagram.

In cases where more than one pure component is present, concentration becomes an important variable. More complex types of phase diagrams, called **binary** and **ternary phase diagrams**, must then be constructed. These phase diagrams have more than two dimensions and can show the effect of more than two variables on the phase of a substance. Such diagrams can also involve substances that take on more than just three states of matter.

The Fe-C and Fe-Cr binary systems and the Fe-Cr-C and Fe-Cr-Ni ternary systems are described below, as an effort to “equip” the reader with the knowledge required to predict each stainless steel type’s microstructure evolution. It should be noted at this early stage though that these phase diagrams can only approximate the actual microstructure that develops in a weld, since: **(1)** stainless steel base and filler metals typically contain up to 10 alloying elements that cannot be accommodated easily with standard phase diagrams, and **(2)** phase diagrams are based on equilibrium conditions, while welding is associated with rapid heating and cooling that result in non-equilibrium conditions.

3.2 IRON-CARBON SYSTEM ^{[2] [6] [7] [10]}

A study of the constitution and structure of all steels must first start with the iron-carbon equilibrium diagram. Many of the basic features of this system influence the behaviour of even the most complex alloy steels.

It should be pointed out right from the beginning that the equilibrium diagram really represents the metastable equilibrium between iron and iron carbide (cementite). Cementite is metastable, and the true equilibrium should be between iron and graphite. Although graphite occurs extensively in cast irons (2wt% to 4wt% C) it is usually difficult to obtain this equilibrium phase in steels (0.03wt% to 1.5wt% C). Therefore, the metastable equilibrium between iron and iron carbide should be considered, since it is relevant to the behaviour of most steels in practice.

The phase field of γ -iron (austenite) is much larger than that of α -iron (ferrite), and that reflects the much greater solubility of carbon in γ -iron, with a maximum value of just over 2wt% at 1147°C (2097°F). This high solubility of carbon in γ -iron is of high importance in heat treatment, when solution treatment in the γ -region followed by rapid quenching in room temperature allows a supersaturated solid solution of carbon to be formed in iron. The α -iron phase is extremely restricted with

only a 0.02wt% carbon solubility at 723°C (1333°F). That means that in steels, whose carbon content ranges from 0.05wt% to 1.5wt%, α -iron is normally associated with iron carbide in one form or another. Similarly the δ -phase field is very restricted between 1390°C (2534°F) and 1594°C (2901°F) and disappears completely when the carbon content reaches 0.5wt%.

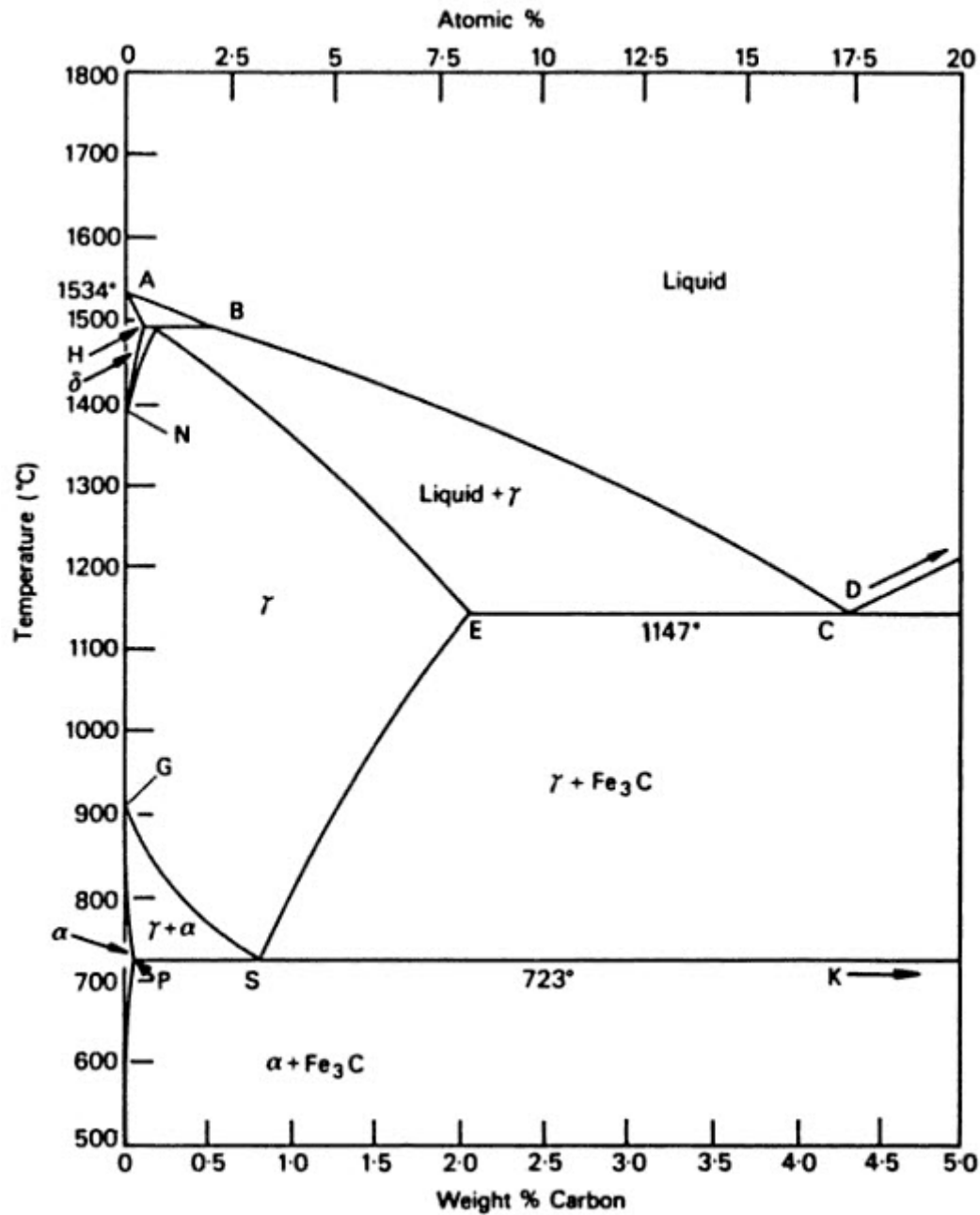


Figure 3.2 A closer look on the Fe-C binary system.

There are several temperatures or critical points in the Fe-C equilibrium diagram which are important, since most of them are also present in stainless steel equilibrium diagrams. Firstly, there is the A_1 temperature (723°C or 1333°F in the binary diagram) at which the eutectoid reaction occurs. Secondly, there is the A_3 temperature when α -iron transforms to γ -iron. For pure iron this occurs at 912°C (1674°F), but the transformation temperature is progressively lowered along the GS line by the addition of carbon. The third point is A_4 at which γ -iron transforms to δ -

iron, 1390°C (2534°F) in pure iron, but this temperature is raised as carbon is added. Lastly, the A_2 point is the Curie point when ferritic iron changes from the ferro- to the paramagnetic condition. This event occurs at 769°C (1416°F) for pure iron, but no change in crystal structure is involved.

The A_1 , A_3 and A_4 points are easily detected by thermal analysis or dilatometry, during cooling or heating cycles, and some hysteresis is observed. Consequently, three values for each point can be obtained, **Ac** for heating (chauffage), **Ar** for cooling (refroidissement) and **Ae** (equilibrium). It should be emphasized though that the Ac and Ar values will be sensitive to the rates of heating and cooling, as well as to the presence of alloying elements.

3.3 IRON-CHROMIUM SYSTEM [2] [6] [7] [8] [10] [11]

The Fe-Cr phase diagram is the starting point for describing stainless steel phase stability, since Cr is the primary alloying element. At first it should be noted that there is complete solubility of Cr in iron at elevated temperatures, and solidification of all Fe-Cr alloys occur as ferrite.

The binary Fe-Cr equilibrium diagram proves that chromium restricts the occurrence of the γ -loop by lowering the A_4 and raising the A_3 transformation temperatures, to the extent that above 12.7wt% Cr the binary alloys are ferrite over the entire temperature range, while there is a narrow $\alpha+\gamma$ range between 12wt% and 13wt% Cr. Alloys with less than about 12wt% Cr will be fully austenitic at temperatures within the γ -loop. Upon rapid cooling the austenite can transform to martensite.

The ferrite is normally referred to as delta ferrite, because in these steels the phase can have a continuous existence from the melting point to room temperature. The addition of carbon to the binary alloy extends the γ -loop to higher chromium contents and also widens the $\alpha+\gamma$ phase field up to 0.3wt%C.

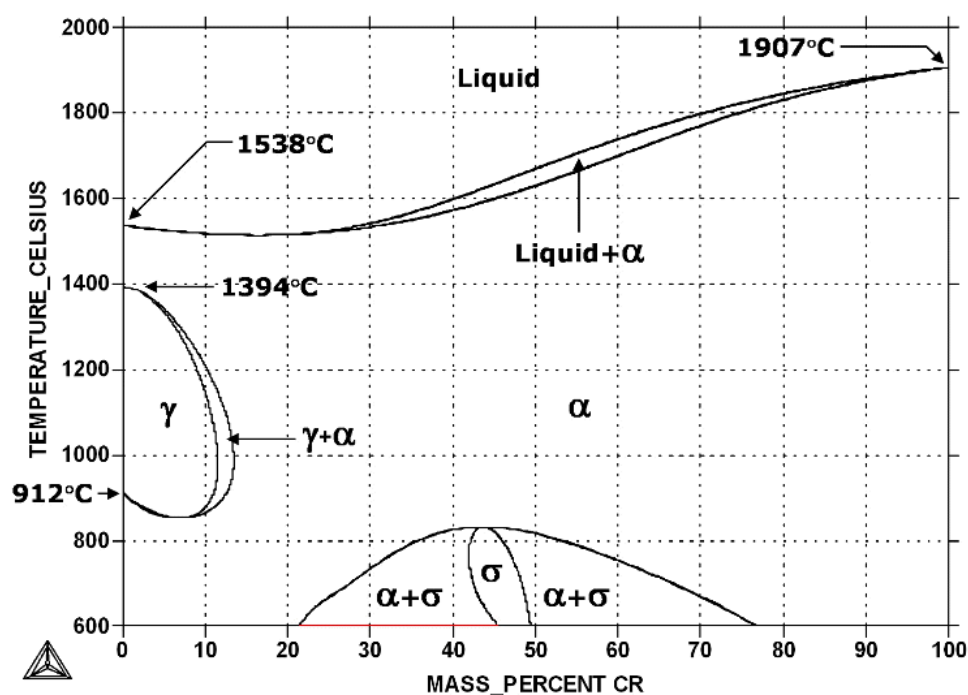


Figure 3.3 A close-up on the Fe-Cr system.

It should also be noted that there is a low-temperature equilibrium phase, called **sigma phase**, present in the Fe-Cr system. This phase is actually an intermetallic Fe-Cr compound that has a tetragonal crystal structure. The sigma phase forms most readily in alloys exceeding 20wt% Cr, such as AISI 316 or AISI 310 stainless steels. Sigma forms at low temperatures. That is why the kinetics of the formation are quite sluggish, and precipitation requires extended time in the temperature range of 600°C to 800°C (1120°F to 1500°F). Sigma is a hard, brittle phase and its harmful effects on mechanical properties and corrosion resistance are well documented. This is why its presence in stainless steels is usually undesirable. Nickel has significant solubility in this phase.

3.4 IRON-CHROMIUM-CARBON SYSTEM ^{[2] [6] [7] [9] [14]}

The Fe-Cr-C system is the basis for both ferritic and martensitic stainless steels.

By adding carbon to the Fe-Cr system we alter the phase equilibrium by significantly complicating it. Carbon is an austenite promoter and as such it expands the γ -loop and allows austenite to be stable at elevated temperatures at much higher Cr contents. It should be noted that even a small amount of carbon can dramatically extend the γ -loop. This is important for development of the martensitic stainless steels, since for martensite to form during cooling, these steels must be austenitic at elevated temperatures. For the ferritic steels on the other hand, the γ -loop has to be controlled so that little or no austenite forms at elevated temperatures.

However, due to the presence of more than two main elements, the tie-lines are no longer contained in an isothermal section, turning the solidification lines into solidification areas and the phase systems is now a three-dimensional system.

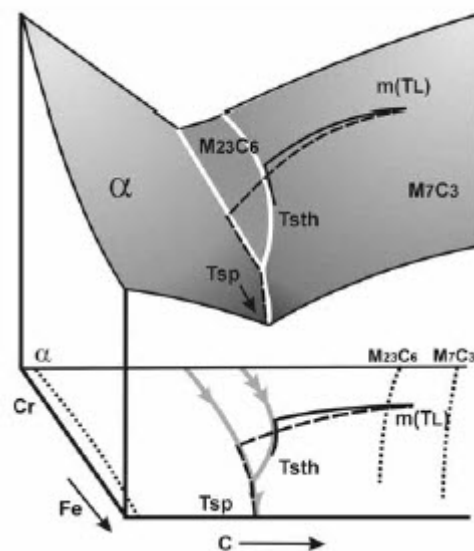


Figure 3.4 The 3-D version of a Fe-Cr-C ternary system.

To allow the Fe-Cr-C ternary system to be viewed as a function of temperature, it is necessary to set one of the elements at a constant value. In this way, a **quasi-binary** or **pseudobinary** phase diagram can be constructed, and its name is derived from the fact that it represents a two-dimensional projection of a three-dimensional system. Because of this, the diagram cannot be used as a binary diagram, and tie-

lines cannot be used to predict the phase balance since they do not necessarily reside on the plane of the diagram. Two pseudobinary diagrams based on the 13wt% Cr and 17wt% Cr systems respectively with variable carbon content are shown below:

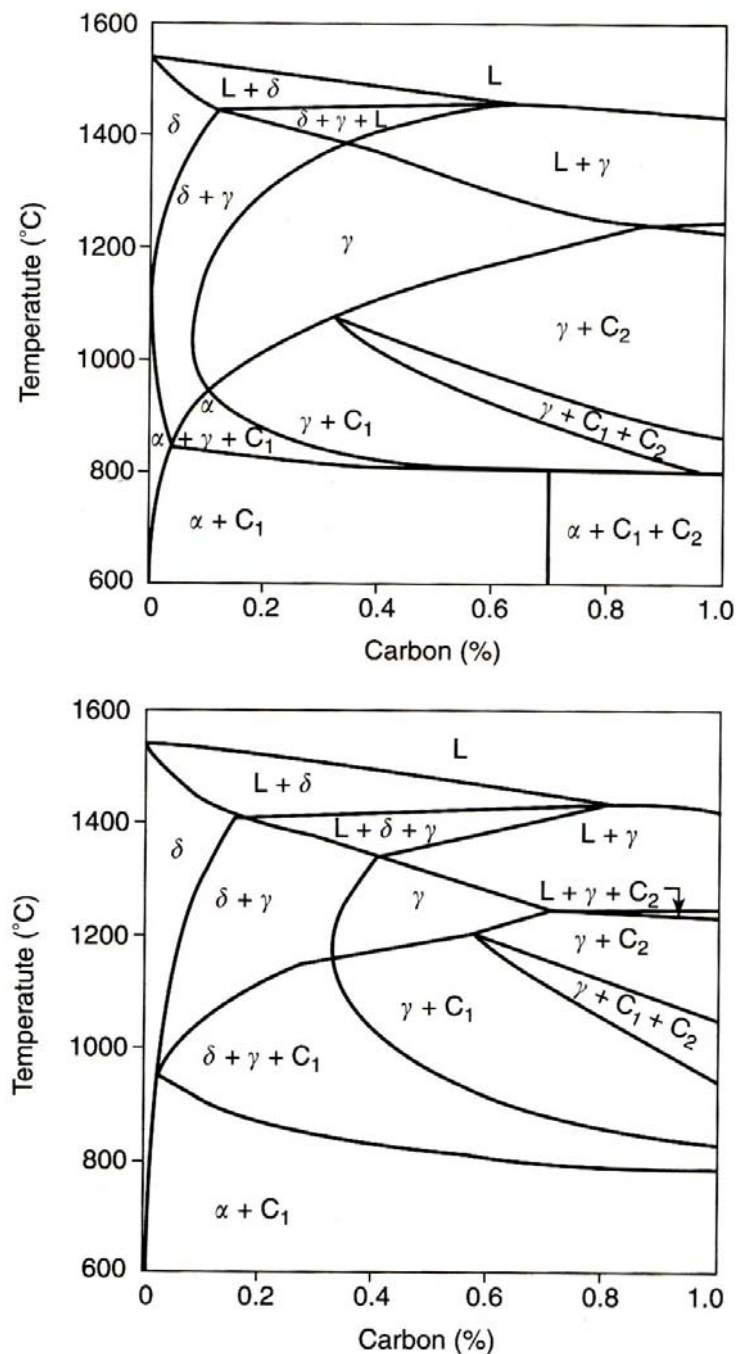


Figure 3.5 Two pseudobinary Fe-Cr-C systems for variable Cr content; 13wt% and 17wt% respectively.

As seen, the addition of carbon had as a result the appearance of two different carbides on the diagram, $(\text{Cr,Fe})_{23}\text{C}_6$ and $(\text{Cr,Fe})_7\text{C}_3$, that led to the presence of additional two- and three-phase fields. The Fe-Cr-C diagram is now obviously much more complex than the Fe-Cr diagram.

The precise knowledge of the limits of the austenite field and of the associated equilibria is of vital importance for steels based on this system, mainly because for compositions lying outside of the austenite field at all temperatures, it is impossible to obtain a martensitic structure by quenching. Furthermore, the type of carbides in equilibrium can have a marked influence on properties, especially on corrosion resistance.

The following diagrams show an example of the limits of the austenite field as mentioned before:

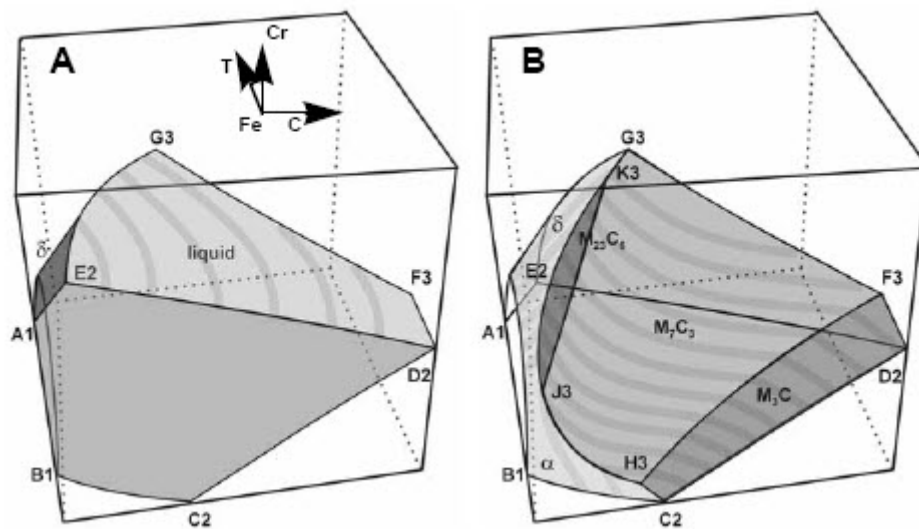


Figure 3.6 Example of the limits of the austenite field and of the associated equilibria.

For low-chromium ferritic and martensitic stainless steels, the 13wt% Cr pseudobinary diagram can be used to explain phase stability and microstructure. At very low carbon contents (less than 0.1wt%) the ternary alloys are fully ferritic at elevated temperatures. If cooled rapidly enough, the alloy will remain primarily ferritic.

At higher carbon contents (above 0.1wt%) austenite will form at elevated temperatures, and mixtures of austenite and ferrite will be present at temperatures just below the solidification temperature range. Upon cooling, the structure will become fully austenitic at temperatures below 1200°C (2200°F). If the cooling rate is rapid enough, this austenite will transform to martensite. This is the base for low-chromium martensitic stainless steels.

At even lower carbon contents (about 0.05wt%) a mixture of austenite and ferrite will be present at elevated temperatures, resulting in a microstructure consisting of ferrite and martensite upon rapid cooling. These microstructures are usually undesirable because of a loss in mechanical properties.

The addition of chromium markedly changes the extent of the austenite field, which disappears completely at about 20wt% Cr. This is illustrated in the following diagram, in which a series of isopleths from 0wt% to 19wt% Cr are superimposed.

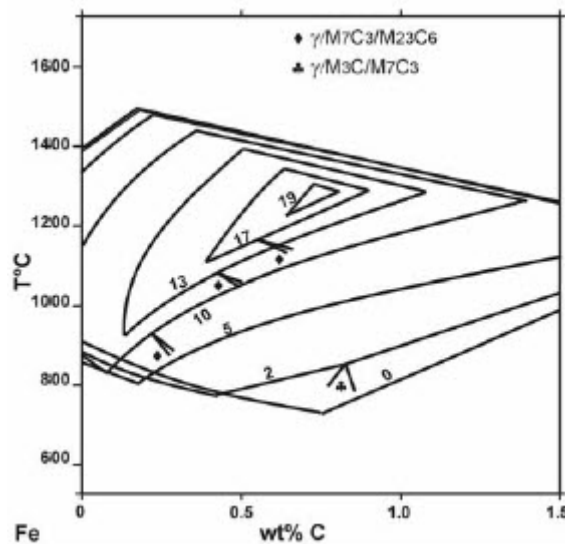


Figure 3.7 Superimposed isopleths of 0wt% to 19wt% Cr content, showing the change in the austenite field.

On the diagram there are also indications of the transition between the $\gamma/\text{M}_3\text{C}/\text{M}_7\text{C}_3$ and $\gamma/\text{M}_7\text{C}_3/\text{M}_{23}\text{C}_6$ three-phase fields.

One other effect of higher chromium content is the expansion of the ferrite phase field, something that is to be expected because of the ferrite-promoting effect of chromium. As a result, the ferrite formed at elevated temperatures will be much more stable, and higher carbon contents are required to form elevated-temperature austenite.

3.5 IRON-CHROMIUM-NICKEL SYSTEM ^{[2] [6] [7] [14] [15]}

The Fe-Cr-Ni system is the basis for both austenitic and duplex stainless steels.

Unlike what happens by adding chromium, addition of nickel to the Fe-Cr system expands the austenite field and allows austenite to be a stable phase at room temperature. Below are two diagrams that represent the liquidus and solidus projections of the Fe-Cr-Ni system respectively.

Upon examination it can be noted that the liquidus surface exhibits a single, dark line that runs from the Fe-rich apex of the triangle to the Cr-Ni side. It separates compositions that solidify as primary ferrite (above and to the left) from those that solidify as primary austenite.

On the other hand, the solidus surface exhibits two dark lines that run from near the Fe-rich apex of the triangle to the Cr-Ni side. Between these two lines the austenite and ferrite phases coexist with liquid just above the solidus, but only with one another just below the solidus. This region separates the ferrite and austenite single-phase fields below the solidus. It should also be noted that these two lines terminate at the ternary eutectic point.

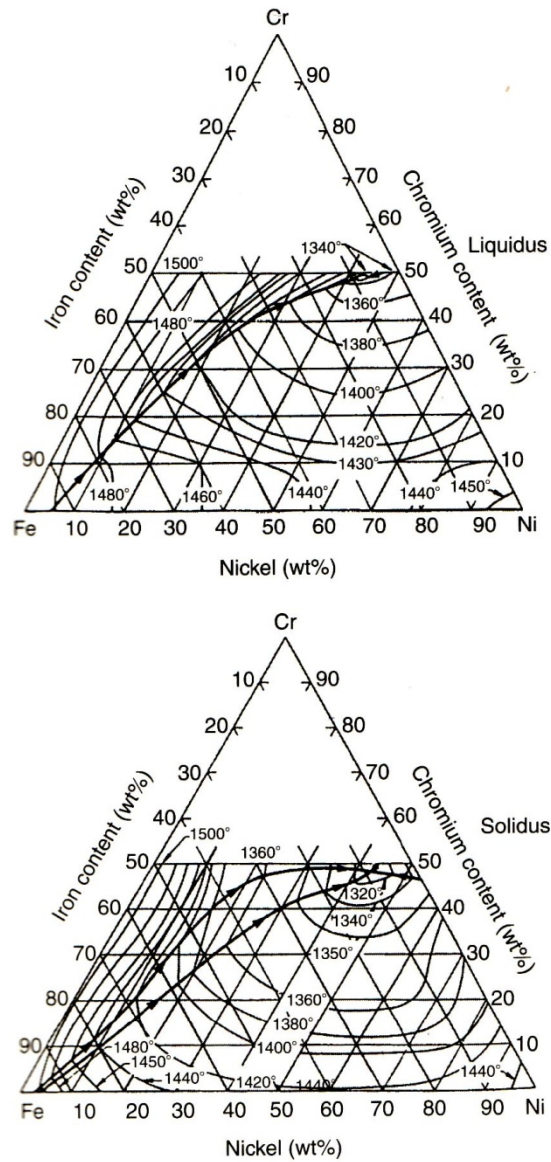


Figure 3.8 Diagrams representing the liquidus and solidus projections of the Fe-Cr-Ni system.

By taking a constant-Fe section through the ternary phase diagram, from the liquidus to room temperature, a pseudobinary Fe-Cr-Ni phase diagram can be generated. Two such diagrams, at 70wt% Fe and at 60wt% Fe, are shown below.

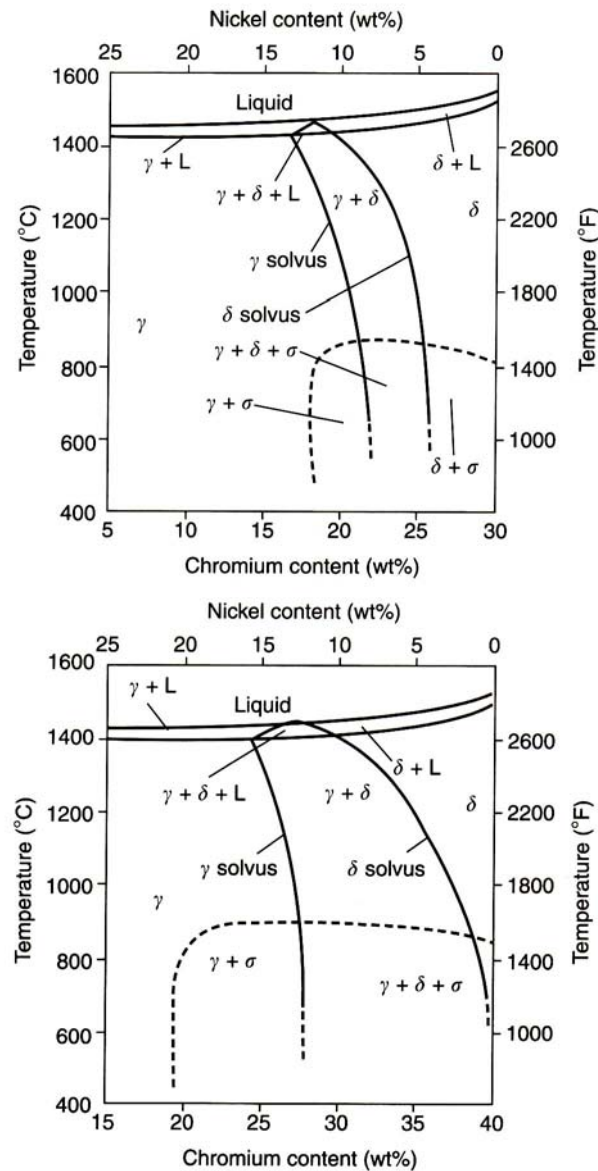


Figure 3.9 Two pseudobinary Fe-Cr-Ni phase diagrams, for 70wt% and 60wt% Fe respectively.

The small triangular region between the solidus and liquidus lines is a three-phase austenite + ferrite + liquid region. Alloys to the left solidify as austenite, while alloys to the right solidify as ferrite. At chromium contents greater than 20wt% and elevated temperature, the ferrite is stable. As the temperature decreases, this ferrite will transform partially to austenite, in the range of 20wt% to 25wt% Cr. Alloys that solidify as austenite (to the left of the aforementioned triangle) remain as austenite upon cooling to room temperature. Alloys that solidify as ferrite (just to the right of the triangle) need to cool through a two-phase austenite + ferrite region. As a result, some of the ferrite transforms into austenite. At compositions further to the right of the triangle ferrite will become increasingly stable.

3.6 CONSTITUTION DIAGRAMS

The prediction of stainless steel weld metal constitution has been a main concern of the scientific community for many years. Most of the research performed until now has been targeted towards the compositional effects on the weld microstructure of these alloys. Various predictive diagrams and equations have been developed and most of them, if not all, are based on the chemical compositions of the alloys of interest. Although the chronological history of every diagram is of no particular interest, the main constitution diagrams that are still in use today will be discussed here.

3.6.1. Schaeffler Diagram ^{[2] [10] [11] [12] [13]}

The prediction of stainless steel weld metal constitution has been the center of interest for many researchers since approximately 1920, when Strauss and Maurer introduced the first nickel-chromium diagram that allowed for the prediction of various phases in the microstructure of wrought, slowly cooled steels. This diagram was modified in 1939 by Scherer et al., who added austenite-ferrite stability lines, and then again by Newell and Fleischman who, recognizing the fact that other elements besides chromium and nickel could have a huge effect on the alloy's microstructure, developed a formula to express the austenite stability on the aforementioned modified diagram.

In 1943, Field et al., based on their study of armor welding, determined that the Newell-Fleischman equation could not be applied directly to weld deposits, since cooling rates associated with welding were much faster than those encountered in wrought products. As a result, they modified the Newell-Fleischman formula as to accommodate for the higher nickel level at which some ferrite can be found.

At the time many investigators had begun to realize that the various alloying elements could be grouped together into equivalency relationships that would incorporate the element's effect on the alloy's microstructure. Campbell and Thomas, Thielemann, Binder et al.; they all made their own contribution on formulae stating these equivalency relationships.

However, it was Anton Schaeffler the first to realize that all preceding research could be put to practical use when combined and applied to welding. Seeing how the Strauss-Maurer diagram could provide a map to predict the microstructures of the wrought chromium-nickel alloys and the Newell-Fleischman, and Field et al. equations could be applied almost directly to welding, Schaeffler directed his research towards combining this information to produce a constitution diagram for weld metals that would allow the prediction of weld metal microstructure based on chemical composition.

Schaeffler incorporated ferrite-promoting elements in a chromium-equivalent formula, and austenite-promoting elements in a nickel-equivalent formula. He then used these formulae as the x- and y-axes respectively and plotted ranges for the specific weld metal microstructures phases in the diagram. The original Schaeffler nickel- and chromium-equivalent equations are given by:

$$Ni_{eq} = wt\%Ni + 0.5 \cdot wt\%Mn + 30 \cdot wt\%C$$

$$Cr_{eq} = wt\%Cr + 2.5 \cdot wt\%Si + 1.8 \cdot wt\%Mo + 2 \cdot wt\%Nb$$

An interesting point of note is that Schaeffler did not include a nitrogen term in his nickel-equivalent formula, despite nitrogen's strong austenite-promoting nature.

This was probably due to the difficulty in determining the nitrogen content of steels in Schaeffler's time, as well as the fact that Schaeffler developed his diagram based on the shielded metal arc welding (SMAW) process, where the nominal nitrogen content was estimated to be about 0.06wt% and thus nitrogen was not considered an alloying element by Schaeffler. However, despite the fact that Schaeffler did not include nitrogen in his nickel-equivalent formula, his diagram proved to be reasonably accurate for most 300-series alloys of his time, using conventional arc welding processes.

Schaeffler further modified his diagram in 1948, turning the austenite/austenite-plus-ferrite boundary into a straight line and adding additional isoferrite lines in the two-phase austenite + ferrite region, while still retaining the original equivalency formulae.

In 1949 Schaeffler eventually introduced the final version of his diagram, which is the version still in use today:

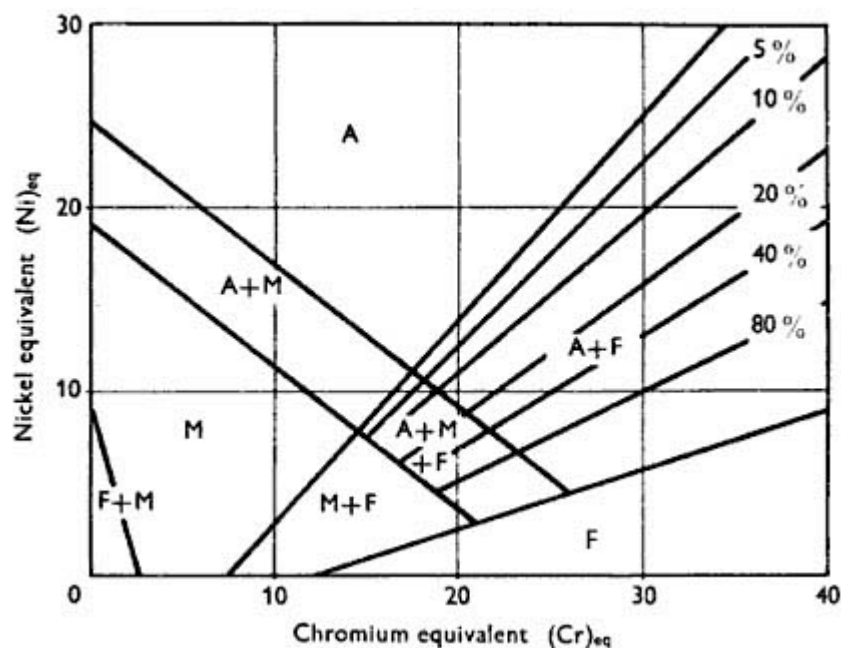


Figure 3.10 The 1949, final version of the Schaeffler diagram.

This diagram also features a new chromium-equivalent formula, since the coefficients for silicon, molybdenum and niobium had been revised. The new equivalency equation was changed to:

$$Cr_{eq} = wt\%Cr + 1.5 \cdot wt\%Si + wt\%Mo + 0.5 \cdot wt\%Nb$$

These equivalency formulae have been changed many times since then, always being empirically determined using the most common austenite- and ferrite-forming elements. In present time, the nickel- and chromium-equivalent formulae are given by the following equations respectively:

$$Ni_{eq} = wt\%Ni + wt\%Co + 0.5 \cdot wt\%Mn + 0.3 \cdot wt\%Cu + 25 \cdot wt\%N + 30 \cdot wt\%C$$

$$Cr_{eq} = wt\%Cr + 2 \cdot wt\%Si + 1.5 \cdot wt\%Mo + 5 \cdot wt\%V + 5.5 \cdot wt\%Al + 1.75 \cdot wt\%Nb + 1.5 \cdot wt\%Ti + 0.75 \cdot wt\%W$$

It should be particularly noted that carbon and nitrogen exert a great deal of influence to the alloy relative to that of the metallic elements, and that becomes apparent by their coefficients within the nickel-equivalent formula.

3.6.2. DeLong Diagram ^{[2] [10] [11] [12] [13]}

The DeLong diagram is, in fact, a refinement of the Schaeffler diagram.

DeLong et al. decided to focus their investigations solely on the 300 series, instead of investigating the entire composition range of stainless steels, and in 1956 they published the first DeLong diagram. The diagram is drawn in an enlarged scale, compared to the Schaeffler diagram, and features more precise lines, thus enabling a more detailed prediction of the ferrite content in austenitic stainless steel weld metal. The DeLong diagram also takes into account the strong austenite stabilizing tendency of nitrogen, and that is depicted in the modified nickel-equivalent equation that is now given by:

$$Ni_{eq} = wt\%Ni + 0.5 \cdot wt\%Mn + 30 \cdot wt\%C + 30 \cdot wt\%N$$

while the chromium-equivalent equation remains unaffected. The addition of the nitrogen term affected the location of the lines in the diagram. Another difference between the Schaeffler and the DeLong diagram is that on the latter the slope of the isoferrite lines was increased to account for any discrepancies discovered between the measured and the calculated ferrite content on high-alloyed stainless steel types. The third difference is that the spacing between isoferrite lines is relatively constant, whereas on the Schaeffler diagram the spacing between isoferrite lines varies.

In 1973 Long and DeLong decided that their initial diagram needed to be further modified. That led to the revised diagram that is still in use and is shown below:

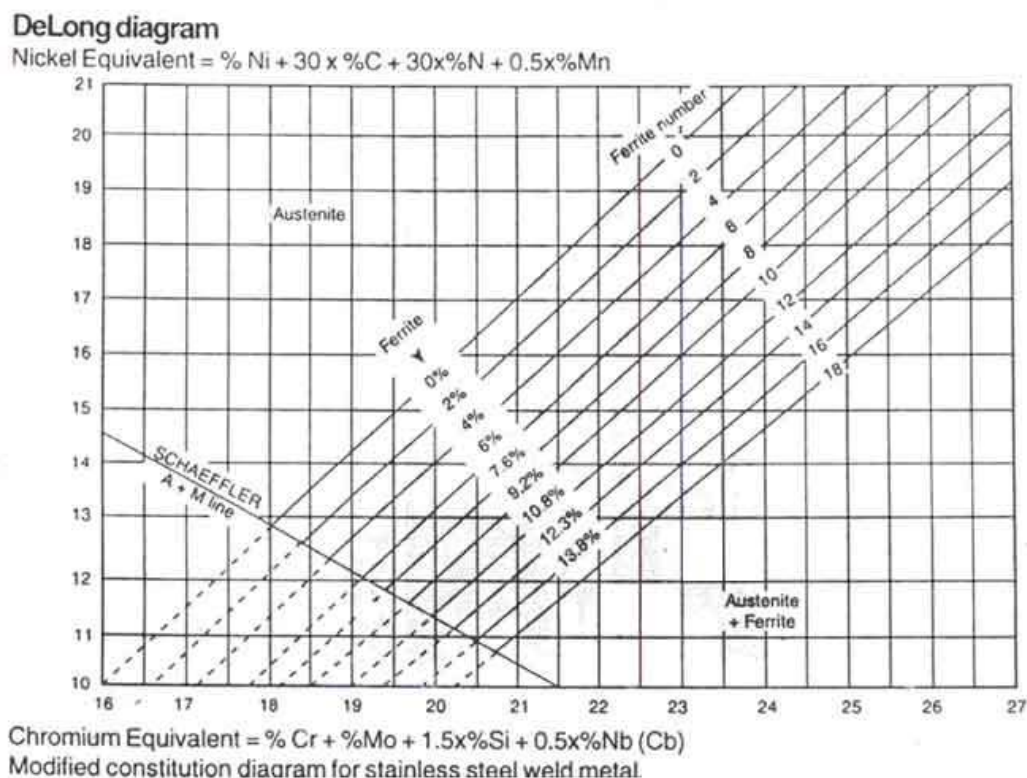


Figure 3.11 The 1973, revised DeLong diagram, containing a Ferrite Number scale.

The major change was the addition of a **Ferrite Number (FN) scale** on the original diagram. The use of FN units was deemed necessary due to the difficulty of measuring the ferrite content quantitatively by volume in stainless steel welds. The FN values are based on magnetic measurements. These measurements are made possible by the fact that the BCC δ -ferrite is ferromagnetic, as opposed to the FCC austenite. One more thing that should be duly noted is that FN values are not intended to relate directly to percent ferrite, although at values below 10 they tend to be similar.

Long and DeLong also reported that their diagram is fairly insensitive to the normal range of heat input variations associated with arc welding. As a result, the diagram can be applied with a fair degree of accuracy to processes such as SMAW, GTAW, GMAW and SAW.

3.6.3. WRC-1992 Diagram^{[2] [10] [11] [12] [13]}

In 1988 WRC funded a study that was conducted by Siewert et al., to revise and expand the Schaeffler and DeLong diagrams. As a consequence, a new diagram was proposed which now covered an expanded range of compositions, ranging from 0 to 100 FN. This diagram became known as the **WRC-1988 diagram**. It was developed by a database of almost 950 welds, gathered from many sources. Its diversity of sources is actually the reason it was considered to be relatively free of any biases in chemical analysis. The WRC-1988 diagram also features a new set of equivalency formulae which removed the manganese coefficient from the nickel equivalent, thereby eliminating the overestimation of FN for highly alloyed weld metals.

After its introduction, the WRC-1988 diagram was extensively evaluated and reviewed by many investigators and its allegedly improved predictive accuracy was tested against data from over 200 welds. At the time, the effect of copper on ferrite content became a topic of interest because of the extensive use of duplex stainless steels, which may contain up to 2wt% copper. As a result, the addition of a coefficient for copper in the WRC-1988 nickel equivalent was deemed necessary to improve the accuracy of FN prediction, and a debate arose within the scientific circles as to which coefficient would be better suited. Finally, it was Kotecki's proposal of a coefficient of 0.25 for copper that formulated the nickel- and chromium-equivalents as:

$$\text{Ni}_{\text{eq}} = \text{wt\%Ni} + 35 \cdot \text{wt\%C} + 20 \cdot \text{wt\%N} + 0.25 \cdot \text{wt\%Cu}$$

$$\text{Cr}_{\text{eq}} = \text{wt\%Cr} + \text{wt\%Mo} + 0.7 \cdot \text{wt\%Nb}$$

and actually led to the introduction of the **WRC-1992 diagram**, which is shown below.

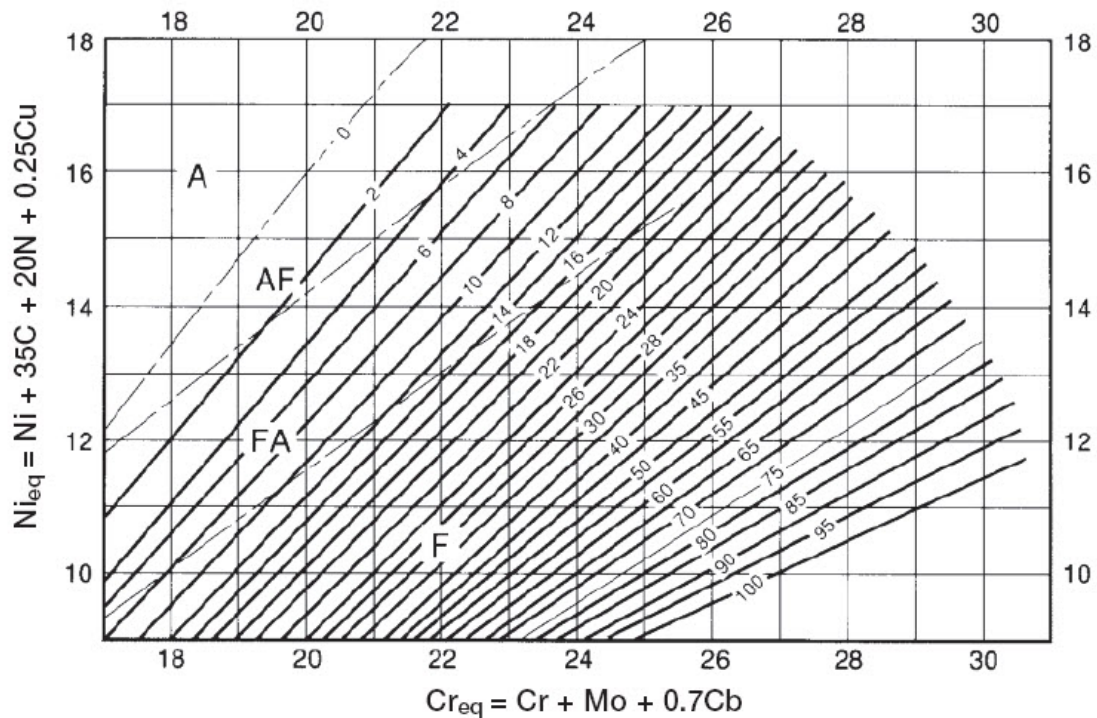


Figure 3.12 The WRC-1992 diagram.

It should be noted, however, that although the extended axes of the diagram allow a wide range of base and filler metal compositions to be plotted, the FN prediction is only valid when the weld metal composition falls within the original iso-FN lines (0 to 100 FN) of the diagram. Extending the iso-FN lines into the surrounding regions could (and most probably would) result in erroneous predictions since the limits of the iso-FN lines were determined by the extent of the original database.

The WRC-1992 diagram is the most reliable and accurate diagram available for prediction of the FN in austenitic and duplex stainless steel weld metals. However, the WRC-1992 diagram is not flawless; one potential shortcoming is the absence of a factor for titanium. Being a potent carbide former, titanium could definitely influence the phase balance by removing carbon from the matrix.

AUSTENITIC STAINLESS STEELS: Fe-Cr-Ni-(Mo) alloys

4.1 GENERAL ^{[2] [16] [25] [26] [27]}

Austenitic stainless steel is dominant in the market and is produced in higher tonnages than any other group of stainless steels, thus representing the largest of the groups of stainless steels. The group exhibits excellent corrosion resistance, ductility and toughness and a rather good weldability.

Austenitic stainless steels have strengths equivalent to those of mild steels with minimum yield strength of approximately 210 MPa at room temperature. They cannot be hardened through heat treatment, but they can be significantly strengthened by cold working. Austenitic steels are characterized by their high content of austenite-formers, especially nickel. They are also alloyed with chromium, molybdenum and sometimes with copper, titanium, niobium and nitrogen. Their chemical composition ($C < 0.08\%$, $16.00\% \leq Cr \leq 28.00\%$, $3.50\% \leq Ni \leq 32.00\%$) allows them to maintain an austenitic structure from cryogenic temperatures up to the alloy's melting point. Because of their high chromium and nickel content, austenitic stainless steels are the most corrosion resistant of the stainless group providing unusually fine mechanical properties.

Of the entire range of austenitic stainless steels, the 300 series is considered to be the oldest and most commonly used. Type 304 is the foundation of this series, and along with 304L, represents the most commonly selected austenitic grade. L grade stainless steels became popular in the 1960's and 1970's with the coming of the AOD melting practice. The use of the AOD melting method reduced the cost differential between standard and L grades.

Austenitic stainless steels are most commonly used for food processing equipment, equipment for the chemical industry, domestic appliances and various architectural applications. This group of stainless steels has limited use in shipbuilding, as described in the appropriate section.

4.2 COMPOSITION AND MECHANICAL PROPERTIES ^{[2] [25] [26] [27] [29] [30]}

Straight grades of austenitic stainless steel contain a maximum of 0.08% carbon. There is a misconception that these grades contain a minimum of 0.03% carbon, but in reality something like that is not required. As long as the material meets the physical requirements of straight grades, there is no minimum carbon required. The most common straight alloys are AISI 304, 316, 321, and 347 and their variants, which are of the “18-8” type. That means that they have nominal values of 18wt% Cr and 8wt% Ni. Types AISI 321 and 347 are also known as “stabilized” grades, since the addition of titanium (for type 321) and tantalum/columbium (for type 347) stabilizes the alloy against the formation of $M_{23}C_6$ chromium carbides. Since these alloying elements form stable MC-type carbides at elevated temperatures, the formation of chromium-rich carbides is restricted, thus reducing the possibility for sensitization that can lead to intergranular corrosion in austenitic stainless steels.

The following figure illustrates how alloying elements can determine the types of the austenitic group:

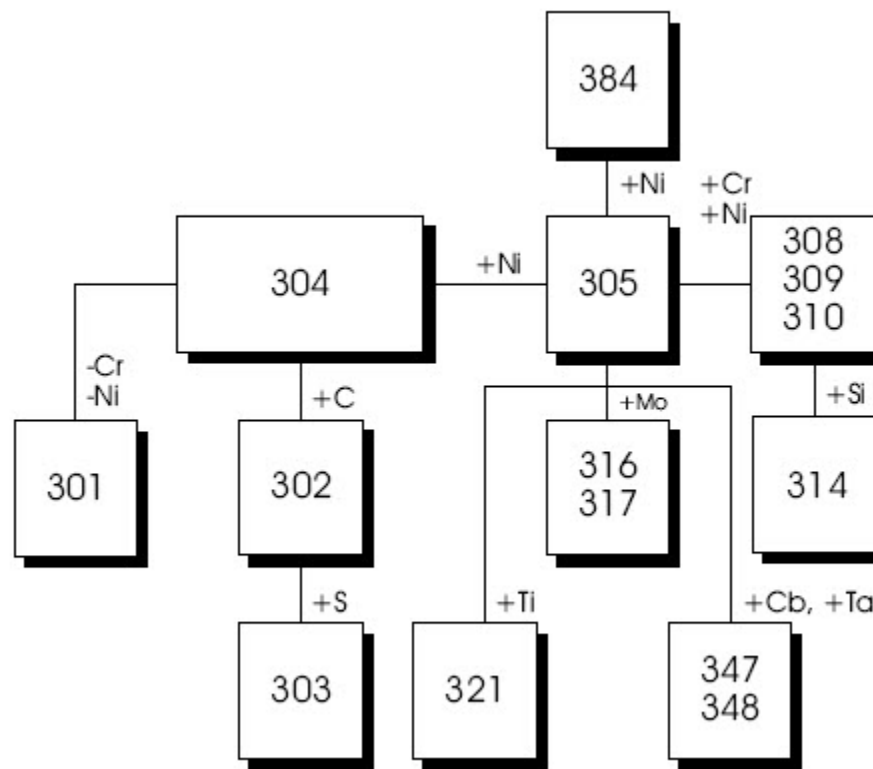


Figure 4.1 The Austenitic Group.

As stated before, the letter “L” after a steel grade indicates low-carbon content. The “L” grades are refined grade steels containing no more than 0.024wt% C. This inhibits carbide precipitation, thus allowing the grade to provide extra corrosion resistance after welding. As described in earlier sections of this thesis, when steel is heated to temperatures in the critical range (800° F to 1600° F or 430° C to 870° C) carbon precipitates out, combines with the chromium and gathers on the grain boundaries. This deprives the steel of the chromium in solution and promotes corrosion adjacent to the grain boundaries. By controlling the amount of carbon this is minimized. But still, not all stainless steels are produced as “L” grades, mainly because “L” grades are more expensive than straight grades. However, “L” grades are used where annealing after welding is impractical, such as in the field where pipe and fittings are being weld.

A grade denoted as “H” contains a minimum of 0.04% carbon and a maximum of 0.10% carbon. Higher carbon concentration enables the material to retain its strength at extreme temperatures. This is the main reason “H” grades are used when the material is expected to operate at extreme temperatures. “H” grades are also known for their improved resistance to intergranular attack in corrosive environments.

Grades denoted as “N” have a minimum requirement of 0.15% nitrogen, thus improving the alloy’s strength, galling resistance, and pitting corrosion. The letter “M” indicates that the alloy contains a minimum of 4.00% molybdenum.

The following table lists a general composition range of Standard Austenitic Stainless Steels:

Chromium:	16 to 25wt%
Nickel:	8 to 20wt%
Manganese:	1 to 2wt%
Silicon:	0.5 to 3wt%
Carbon:	0.02 to 0.08wt% (less than 0.03wt% for “L” grades)
Molybdenum:	0 to 2wt%
Nitrogen:	0 to 0.15wt%
Titanium and Niobium:	0 to 0.2wt%

More detailed compositions of Standard Wrought Austenitic Stainless Steels are shown on the following table:

Type	UNS Number	Composition - Percent *							Other
		C	Mn	Si	Cr	Ni	P	S	
201	S20100	0.15	5.5-7.5	1.00	16.0-18.0	3.5-5.5	0.06	0.03	0.25 N
202	S20200	0.15	7.5-10.0	1.00	17.0-19.0	4.0-6.0	0.06	0.03	0.25 N
205	S20500	0.12-0.25	14.0-15.5	1.00	16.5-18.0	1.0-1.75	0.06	0.03	0.32-0.40 N
216	S21600	0.08	7.5-9.0	1.00	17.5-22.0	5.0-7.0	0.045	0.03	2.0-3.0 Mo; 0.25-0.5 N
301	S30100	0.15	2.00	1.00	16.0-18.0	6.0-8.0	0.045	0.03	
302	S30200	0.15	2.00	1.00	17.0-19.0	8.0-10.0	0.045	0.03	
302B	S30215	0.15	2.00	2.0-3.0	17.0-19.0	8.0-10.0	0.045	0.03	
303**	S30300	0.15	2.00	1.00	17.0-19.0	8.0-10.0	0.20	0.15 min.	0.6 Mo
303Se**	S30323	0.15	2.00	1.00	17.0-19.0	8.0-10.0	0.20	0.06	0.15 min. Se
304	S30400	0.08	2.00	1.00	18.0-20.0	8.0-10.5	0.045	0.03	
304H	S30409	0.04-0.10	2.00	1.00	18.0-20.0	8.0-10.5	0.045	0.03	
304L	S30403	0.03	2.00	1.00	18.0-20.0	8.0-12.0	0.045	0.03	
304LN	S30453	0.03	2.00	1.00	18.0-20.0	8.0-10.5	0.045	0.03	0.10-0.15 N
S30430	S30430	0.08	2.00	1.00	17.0-19.0	8.0-10.0	0.045	0.03	3.0-4.0 Cu
304N	S30451	0.08	2.00	1.00	18.0-20.0	8.0-10.5	0.045	0.03	0.10-0.16 N
304HN	S30452	0.04-0.10	2.00	1.00	18.0-20.0	8.0-10.5	0.045	0.03	0.10-0.16 N
305	S30500	0.12	2.00	1.00	17.0-19.0	10.5-13.0	0.045	0.03	
308	S30800	0.08	2.00	1.00	19.0-21.0	10.0-12.0	0.045	0.03	
308L		0.03	2.00	1.00	19.0-21.0	10.0-12.0	0.045	0.03	
309	S30900	0.20	2.00	1.00	22.0-24.0	12.0-15.0	0.045	0.03	
309S	S30908	0.08	2.00	1.00	22.0-24.0	12.0-15.0	0.045	0.03	
309S Cb	S30940	0.08	2.00	1.00	22.0-24.0	12.0-15.0	0.045	0.03	8 x %C - Nb(Cb)
309 Cb + Ta		0.08	2.00	1.00	22.0-24.0	12.0-15.0	0.045	0.03	8 x %C (Nb(Cb) + Ta)
310	S31000	0.25	2.00	1.50	24.0-26.0	19.0-22.0	0.045	0.03	
310S	S31008	0.08	2.00	1.50	24.0-26.0	19.0-22.0	0.045	0.03	
312		0.15	2.00	1.00	30.0 nom.	9.0 nom.	0.045	0.03	
254SMo	S31254	0.020	1.00	0.80	19.5-20.5	17.50-18.5	0.03	0.010	6.00-6.50Mo; 0.18-0.22N; Cu=0.5-1.00
314	S31400	0.25	2.00	1.5-3.0	23.0-26.0	19.0-22.0	0.045	0.03	
316	S31600	0.08	2.00	1.00	16.0-18.0	10.0-14.0	0.045	0.03	2.0-3.0 Mo
316F**	S31620	0.08	2.00	1.00	16.0-18.0	10.0-14.0	0.20	0.10 min.	1.75-2.5 Mo
316H	S31609	0.04-0.10	2.00	1.00	16.0-18.0	10.0-14.0	0.045	0.03	2.0-3.0 Mo
316L	S31603	0.03	2.00	1.00	16.0-18.0	10.0-14.0	0.045	0.03	2.0-3.0 Mo
316LN	S31653	0.03	2.00	1.00	16.0-18.0	10.0-14.0	0.045	0.03	2.0-3.0 Mo; 0.10-0.30 N
316N	S31651	0.08	2.00	1.00	16.0-18.0	10.0-14.0	0.045	0.03	2.0-3.0 Mo; 0.10-0.16 N
317	S31700	0.08	2.00	1.00	18.0-20.0	11.0-15.0	0.045	0.03	3.0-4.0 Mo
317L	S31703	0.03	2.00	1.00	18.0-20.0	11.0-15.0	0.045	0.03	3.0-4.0 Mo
317M	S31725	0.03	2.00	1.00	18.0-20.0	12.0-16.0	0.045	0.03	4.0-5.0 Mo
321	S32100	0.08	2.00	1.00	17.0-19.0	9.0-12.0	0.045	0.03	5 x %C min. Ti
321H	S32109	0.04-0.10	2.00	1.00	17.0-19.0	9.0-12.0	0.045	0.03	5 x %C min. Ti
329	S32900	0.10	2.00	1.00	25.0-30.0	3.0-6.0	0.045	0.03	1.0-2.0 Mo
330	N08930	0.08	2.00	0.75-1.5	17.0-20.0	34.0-37.0	0.04	0.03	
AL6-XN	N80367	0.030	2.00	1.00	20.0-22.0	23.5-25.5	0.04	0.03	6.00-7.00Mo; 0.18-0.25N; Cu=0.75
330HC		0.40	1.50	1.25	19.0 nom.	35.0 nom.			
332		0.04	1.00	0.50	21.5 nom.	32.0 nom.	0.045	0.03	
347	S34700	0.08	2.00	1.00	17.0-19.0	9.0-13.0	0.045	0.03	10 x %C min. Nb(Cb) + Ta
347H	S34709	0.04-0.10	2.00	1.00	17.0-19.0	9.0-13.0	0.045	0.03	10 x %C min. Nb(Cb) + Ta
348	S34800	0.08	2.00	1.00	17.0-19.0	9.0-13.0	0.045	0.03	0.2 Cu; 10 x %C min. Nb(Cb) + Ta(c)
348H	S34809	0.04-0.10	2.00	1.00	17.0-19.0	9.0-13.0	0.045	0.03	0.2 Cu; 10 x %C min. Nb(Cb) + Ta
384	S38400	0.08	2.00	1.00	15.0-17.0	17.0-19.0	0.045	0.03	
Nitronic 32	S24100	0.10	12.0	0.50	18.0	1.6			0.35 N
Nitronic 33	S24000	0.06	13.0	0.5	18.0	3.0			0.30 N
Nitronic 40	S21900	0.08	8.0-10.0	1.00	18.0-20.0	5.0-7.0	0.06	0.03	0.15-0.40 N
Nitronic 50	S20910	0.06	4.0-6.0	1.00	20.5-23.5	11.5-13.5	0.04	0.03	1.5-3.0 Mo; 0.2-0.4 N; 0.1-0.3 Cb; 0.1-0.3 V
Nitronic 60	S21800	0.10	7.0-9.0	3.5-4.5	16.0-18.0	8.0-9.0	0.04	0.03	1.5-3.0 Mo; 0.2-0.4 N;

Figure 4.2 Austenitic stainless steels composition table.

The mechanical properties, corrosion resistance and weldability of steels are largely determined by its microstructure. The presence of ferrite in the wrought microstructure

can reduce the ductility and, potentially, the toughness of austenitic stainless steels. Ferritic regions in steels can also be preferential sites for the precipitation of $M_{23}C_6$ carbides and sigma phase. Sigma phase is a potent embrittling agent when it comes to stainless steels. On the other hand, $M_{23}C_6$ carbide precipitation has a detrimental effect on corrosion resistance, since the presence of these grain boundary carbides can lead to intergranular corrosion when exposed to certain environments. Unfortunately, this precipitation reaction is accelerated in alloys that are strengthened by cold work.

In more detail, alloying with nitrogen raises the steels' yield strength. The molybdenum-free steels also have very good high-temperature properties and exhibit good impact strength at low temperature too.

As one can clearly see in the following stress-strain curve, austenitic steels exhibit low yield strength and excellent ductility. In truth, austenitic steels have somewhat lower yield strength than ferritic steels and remarkably lower yield strength than ferritic-austenitic (duplex) steels. Their ductility however is unsurpassable by any of the other grades, allowing austenitic grades to exhibit very high elongation and be very tough.

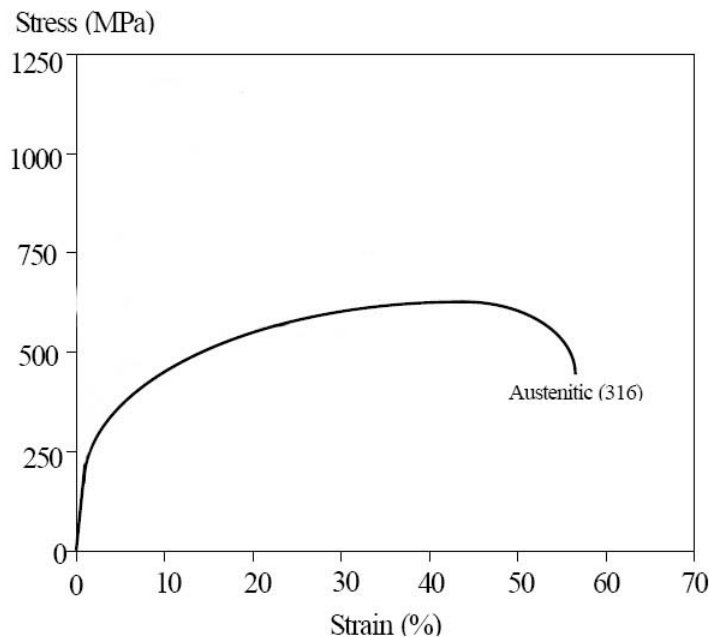


Figure 4.3 Stress-strain curve for the Type 316 austenitic stainless steel.

Austenitic stainless steels are characterized by strong work hardening and their strength increases with increasing levels of nitrogen, molybdenum and carbon, even though carbon is never used to achieve increased strength due to its negative effect on corrosion resistance.

The toughness of the different types of stainless steel shows considerable variation. Toughness is dependent on temperature and, as a general rule, it increases with increasing temperature. One measure of toughness is the impact toughness and following diagram makes clear that austenitic steels do not exhibit any toughness transition as the other steel types (as will be described in later chapters of this thesis),

but instead have excellent toughness at all temperatures. Austenitic steels are thus preferable for low temperature applications.

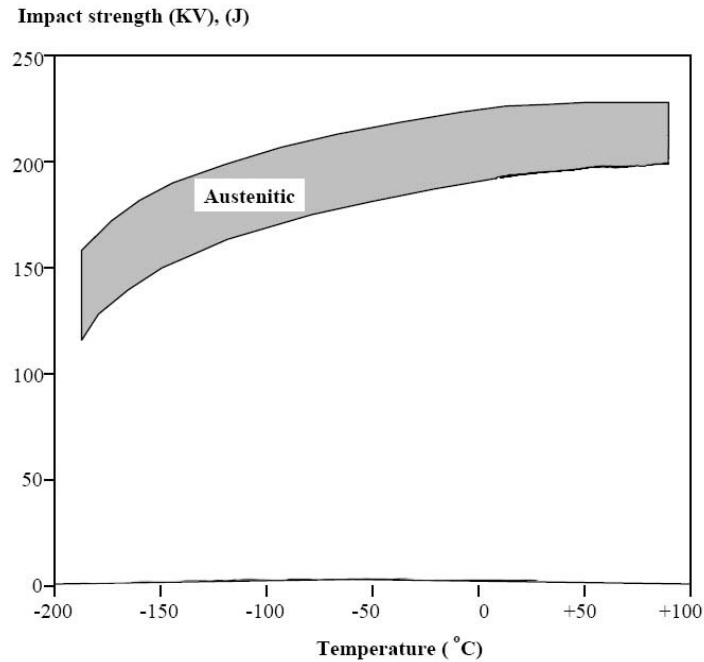


Figure 4.4 Impact toughness for austenitic stainless steels.

During cyclic loading stainless steels will fail at stress levels considerably lower than the tensile strength measured during tensile testing. Each material can withstand a certain number of load cycles and that number depends on the stress amplitude. The number of cycles until failure increases with decreasing load amplitude, until a certain amplitude is reached, below which no failure occurs. This is the material's fatigue limit. However, sometimes there is no fatigue limit, but the stress amplitude shows a slow decrease with increasing number of cycles. In such cases the fatigue strength is given in relation to a certain number of cycles. The following diagram shows the S-N (Wöhler) curve for an austenitic stainless steel of Type 316(hMo) in air.

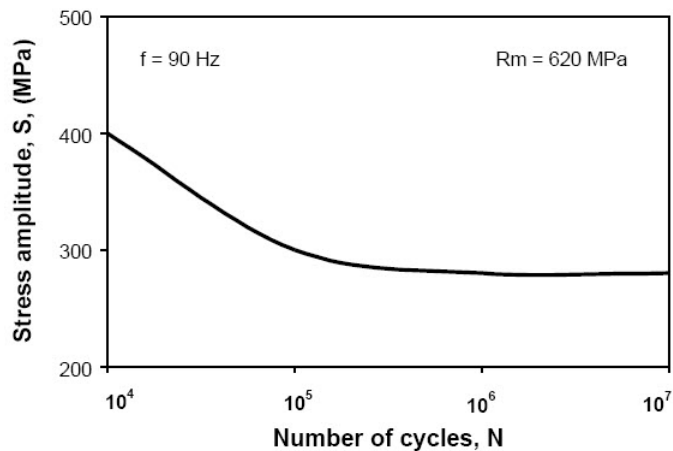


Figure 4.5 A fatigue limit diagram for austenitic stainless steels.

The fatigue properties, described by the Wöhler curve with a fatigue limit at a lifetime of 10^6 - 10^7 load cycles, austenitic steels can be related to their tensile strength as shown in the following table, where the relation between the fatigue limit and the tensile strength is also dependent on the type of load or stress ratio. The stress ratio is ratio of the minimum stress to the maximum stress during the load cycle (compressive stresses are defined as negative).

Steel category	S_o/R_m Stress Ratio		Maximum stress
	R = -1	R = 0	
Austenitic	0.45	0.3	yield strength

The service environment of the material has a great impact to fatigue strength. Under both cycling loading and corrosive conditions (also known as “corrosive fatigue”) the fatigue strength will generally decrease. The more aggressive the corrosive environment and the lower the loading frequency, the more potent the effect of the environment. When the loading frequency is very high the corrosion has very little time to make an impact and the fatigue properties of the material will mostly determine its service life. At lower frequencies the corrosion is more pronounced. In such cases an aggressive environment may also cause corrosion attacks that will act as stress concentrations and thus actively shorten the material’s service life.

The following figure illustrates the high temperature strength of austenitic stainless steels, where the dashed line shows the elevated temperature strength of some very high alloyed and nitrogen alloyed austenitic steels:

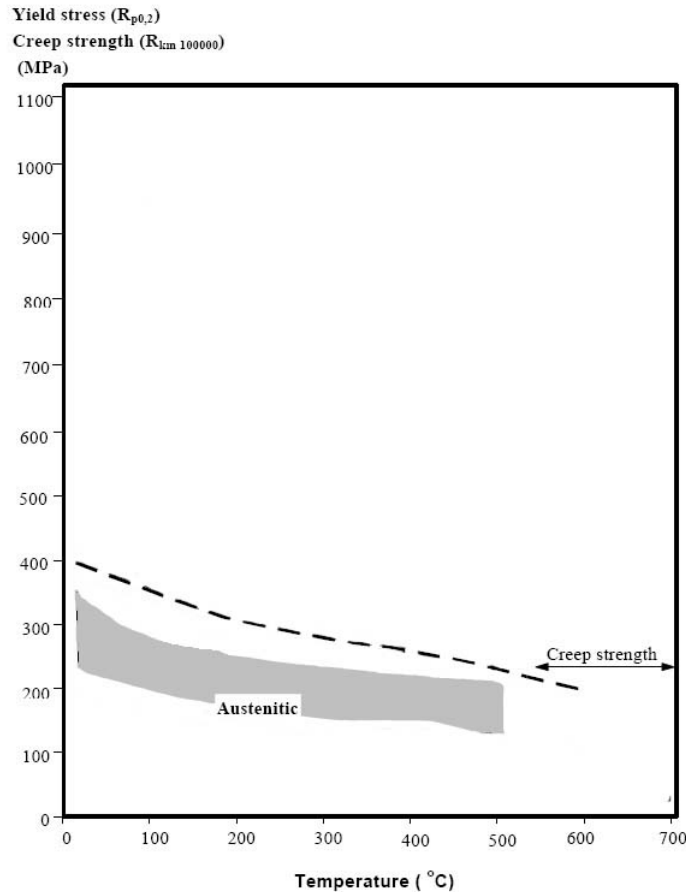


Figure 4.6 A high temperature strength diagram for austenitic stainless steels.

In the temperature range up to 500° C (932° F) most austenitic steels tend to exhibit lower strength than other types of stainless steels, with nitrogen alloyed grades and those containing titanium and niobium being the austenitic steels to exhibit the highest elevated temperature strength. As shown in the figure above the elevated temperature strengths of most austenitic steels fall within the marked area. However, in terms of creep strength the austenitic stainless steels are superior to all other types of stainless steel.

The following table is derived from **ASM Metals Handbook, 8th edition, volume 1**, **ASM Metals Handbook, 9th edition, volume 3** and **ASTM standards** and synthesizes the mechanical properties of austenitic stainless steels:

4: AUSTENITIC STAINLESS STEELS

Type	Condition	Tensile Strength		0.2% Yield Strength		Elong.	R.A.	Hardness
		Ksi	MPa	Ksi	MPa	%	%	Rockwell
201	Anneal	115	793	55	379	55		B90
201	Full Hard	185	1275	140	965	4		C41
202	Anneal	105	724	55	379	55		B90
301	Anneal	110	758	40	276	60		B85
301	Full Hard	185	1275	140	965	8		C41
302	Anneal	90	620	37	255	55	65	B82
302B	Anneal	95	655	40	276	50	65	B85
303	Anneal	90	620	35	241	50	55	B84
304	Anneal	85	586	35	241	55	65	B80
304L	Anneal	80	552	30	207	55	65	B76
304N	Anneal	85	586	35	241	30		
304LN	Anneal	80	552	30	207			
305	Anneal	85	586	37	255	55	70	B82
308	Anneal	85	586	35	241	55	65	B80
308L	Anneal	80	551	30	207	55	65	B76
309	Anneal	90	620	40	276	45	65	B85
310	Anneal	95	655	40	276	45	65	B87
312	Anneal	95	655			20		
314	Anneal	100	689	50	345	45	60	B87
316	Anneal	85	586	35	241	55	70	B80
316L	Anneal	78	538	30	207	55	65	B76
316F	Anneal	85	586	35	241	55	70	B80
317	Anneal	90	620	40	276	50	55	B85
317L	Anneal	85	586	35	241	50	55	B80
321	Anneal	87	599	35	241	55	65	B80
347/348	Anneal	92	634	35	241	50	65	B84
329	Anneal	105	724	80	552	25	50	B98
330	Anneal	80	550	35	241	30		B80
330HC	Anneal	85	586	42	290	45	65	
332	Anneal	80	552	35	241	45	70	
384	Anneal	80	550					

Figure 4.7 The mechanical properties of austenitic stainless steels.

4.3 PHYSICAL PROPERTIES^{[1] [2] [25] [29] [30] [31] [32] [37]}

4.3.1. General^[37]

GRADES	Physical properties of stainless steels									
	DESIGNATIONS (▼)		DENSITY	MODULUS OF ELASTICITY	MEAN COEFFICIENT OF THERMAL EXPANSION		THERMAL CONDUCTIVITY	SPECIFIC HEAT	ELECTRICAL RESISTIVITY	
	EN [N°]	AISI/ASTM	at 20°C	at 20°C	[10 ⁻⁶ °K ⁻¹]		at 20°C	at 20°C	at 20°C	
			[kg/dm ³]	[GPa]	20°C + 200°C	20°C + 400°C	[W/(m·K)]	[J/(kg·K)]	[Ω·mm ² /m]	
AUSTENITIC	1.4372 ^(a)	201	7,8	200	15,7 ^(a)	17,5 ^(b)	15	500 ^(a)	0,70	NON MAGNETIC
	1.4373 ^(a)	202	7,8	200	17,5 ^(f)	18,4 ^(b)	15	503 ^(a)	0,70	
	1.4371 ^(a)		7,8	200	17,5	18,5	15	500	0,70	
	1.4597 ^(a)		7,8	200	16,5	17,0	15	500	0,73	
	1.4369 ^(a)		7,9	190	17,0	18,5	15	500	0,70	
	1.4310 ^(a)	301	7,9	200	17,0	18,0	15	500	0,73	
	1.4319 ^(a)		7,9	200	16,5	17,5	15	500	0,73	
	1.4318 ^(a)	301LN (301L)	7,9	200	16,5	17,5	15	500	0,73	
		302 ^(b)	8,06	193	17,2 ^(a)	17,8 ^(b)	16,3 ^(c)	503	0,72	
	1.4305 ^(a)	303	7,9	200	16,5	17,5	15	500	0,73	
	1.4301 ^(a)	304	7,9	200	16,5	17,5	15	500	0,73	
	1.4311 ^(a)	304LN	7,9	200	16,5	17,5	15	500	0,73	
	1.4948 ^(a)	304H	7,9	200	16,9	17,8	17	450	0,71	
	1.4307 ^(a)	304L	7,9	200	16,5	18,0	15	500	0,73	
	1.4306 ^(a)	304L	7,9	200	16,5	17,5	15	500	0,73	
	1.4315 ^(a)	304N	7,9	200	16,5	17,5	15	500	0,73	
	1.4303 ^(a)	305	7,9	200	16,5	17,5	15	500	0,73	
	1.4567 ^(a)		7,9	200	17,2	18,1	11,3 ^(h)	503 ⁽ⁱ⁾	0,72 ^(a)	
		308 ^(b)	8,06	193	17,3 ^(a)	17,8 ^(b)	15,3 ^(c)	503	0,72	
	1.4828 ^(a)		7,9	200 ^(a)	16,5	17,5	15	500	0,85	
1.4833 ^(a)	309S	7,9	200 ^(a)	16,0	17,5	15	500	0,78		
	310 ^(b)	8,06	200	15,9 ^(a)	16,2 ^(b)	14,2 ^(c)	503	0,78		
1.4845 ^(a)	310S	7,9	200 ^(a)	15,5	17,0	15	500	0,85		
1.4841 ^(a)	314	7,9	200 ^(a)	15,5	17,0	15	500	0,90		
AUSTENITIC	1.4401 ^(a)	316	8,0	200	16,5	17,5	15	500	0,75	NON MAGNETIC
	1.4436 ^(a)	316	8,0	200	16,5	17,5	15	500	0,75	
		316F ^(b)	8,06	200	16,5 ^(a)	17,5 ^(b)	14,4 ^(c)	486	0,74	
		316N ^(b)	8,06	196	(n.r.)	(n.r.)	(n.r.)	503	0,74	
		316H ^(b)	8,03	193	15,9 ^(a)	16,2 ^(b)	16,3 ^(c)	500	0,74	
	1.4404 ^(a)	316L	8,0	200	16,5	17,5	15	500	0,75	
	1.4435 ^(a)	316L	8,0	200	16,5	17,5	15	500	0,75	
	1.4432 ^(a)	316L	8,0	200	16,5	17,5	15	500	0,75	
	1.4406 ^(a)	316LN	8,0	200	16,5	17,5	15	500	0,75	
	1.4429 ^(a)	316LN	8,0	200	16,5	17,5	15	500	0,75	
	1.4571 ^(a)	316Ti	8,0	200	17,5	18,5	15	500	0,75	
	1.4580 ^(a)	316Cb	8,0	200	17,5	18,5	15	500	0,75	
		317 ^(b)	8,06	193	16,0 ^(a)	16,2 ^(b)	16,3 ^(c)	503	0,74	
	1.4438 ^(a)	317L	8,0	200	16,5	17,5	14	500	0,85	
	1.4434 ^(a)	317LN	8,0	200	16,5	17,5	15	500	0,75	
	1.4439 ^(a)	317 LMN	8,0	200	16,5	17,5	14	500	0,85	
	1.4541 ^(a)	321	7,9	200	16,5	17,5	15	500	0,73	
	1.4878 ^(a)	321H	7,9	200 ^(a)	17,0	18,0	15	500	0,73	
	1.4550 ^(a)	347	7,9	200	16,5	17,5	15	500	0,73	
		347H ^(b)	8,03	193	16,6 ^(a)	18,2 ^(b)	16,1 ^(c)	500	0,72	
1.4335 ^(a)		7,9	195	16,1	16,9	14	450	0,85		
1.4466 ^(a)	310MoLN	8,0	195	15,7 ^(a)	17,0 ^(b)	14	500	0,80		
1.4361 ^(a)		7,7	200	(n.r.)	(n.r.)	14	(n.r.)	(n.r.)		
1.4570 ^(a)		7,9	200	16,5	17,5	15	500	0,70		

AUSTENITIC	1.4578 ^(a)		8,0	200	(n.r.)	(n.r.)	(n.r.)	(n.r.)	(n.r.)	NON MAGNETIC
	1.4563 ^(a)		8,0	195	16,1	16,9	12	450	1,00	
	1.4537 ^(a)		8,1	195	(n.r.)	(n.r.)	14	500	0,85	
	1.4539 ^(a)	904L	8,0	195	16,1	16,9	12	450	1,00	
	1.4547 ^(a)		8,0	195	17,0	18,0	14	500	0,85	
	1.4652 ^(a)		8,0	190	15,4	16,2	8,6	500	0,78	
	1.4659 ^(a)		8,2	190	15,5	16,3	12	450	1,00	
	1.4529 ^(a)		8,1	195	16,1	16,9	12	450	1,00	
	1.4565 ^(a)		8,0	190	15,5	16,8	12	450	0,92	
	1.4664 ^(a)	330	8,0	196 ^(b)	15,0	16,0	12,5	550	1,00	
	1.4635 ^(a)		7,8	(n.r.)	17,0	18,0	15	500	0,85	
	1.4676 ^(a)		8,0	196 ^(b)	15,0	16,0	12	550	1,00	
	1.4677 ^(a)		8,0	(n.r.)	15,5	16,5	12	450	0,96	
	1.4618 ^(a)		7,8	(n.r.)	16,5	18,0	15	500	0,85	
	1.4654 ^(a)		7,9	(n.r.)	15,5	16,5	11	450	1,00	
		384 ^(b)	8,06	193	17,2 ^(b)	(n.r.)	16,3 ^(b)	503	0,79	

Figure 4.8 The physical properties of austenitic stainless steels.

4.3.2. Weldability and Issues ^{[2] [25] [30] [31] [32]}

Methods commonly used with austenitic stainless steels are:

- Shielded Meta Arc Welding (SMAW) with covered electrodes
- Gas Tungsten Arc Welding TIG (GTAW) with pure argon as the shielding gas
- Gas Metal Arc Welding MIG (GMAW) with pure argon as the shielding gas
- Plasma Arc Welding (PAW) with pure argon as the shielding gas
- Submerged Arc Welding (SAW)
- Stud Welding (SW)

All stainless steels solidify with either austenite or ferrite as their primary phase. Depending on their composition, however, austenitic stainless steels may solidify as primary austenite or primary ferrite. Small changes in their composition may promote a shift from primary ferrite to primary austenite, but the composition of many austenitic alloys is broad enough to allow both solidification modes. As such, there are four solidification and solid-state transformation possibilities for austenitic stainless steel weld metals: Type A, Type AF, Type FA, and Type F. The reactions leading to these types are listed on the following table:

Solidification Types, Reactions, and Resultant Microstructures		
Solidification Type	Reaction	Microstructure
A	$L \rightarrow L + A \rightarrow A$	Fully austenitic, well-defined solidification structure
AF	$L \rightarrow L + A \rightarrow L + A + (A + F)_{eut} \rightarrow A + F_{eut}$	Ferrite at cell and dendrite boundaries
FA	$L \rightarrow L + F \rightarrow L + F + (F + A)_{per/eut} \rightarrow F + A$	Skeletal and/or lathy ferrite resulting from ferrite-to-austenite transformation
F	$L \rightarrow L + F \rightarrow F \rightarrow F + A$	Acicular ferrite or ferrite matrix with grain boundary austenite and Widmanstätten side plates

Figure 4.9 Solidification and solid-state transformation possibilities for austenitic stainless steels.

Note that Types A and AF are associated with primary austenite solidification, where austenite is the first phase to form upon solidification, while Types FA and F have delta ferrite as their primary phase. These transformations are very important, since they play a huge role on the various boundaries that are present in austenitic stainless steel weld metals. Many of the defects associated with the fusion zone both during fabrication and in service are also associated with these boundaries; that alone underlines the importance of understanding these boundaries.

As shown on the following figure, there are at least three different types of boundaries that can be observed metallographically:

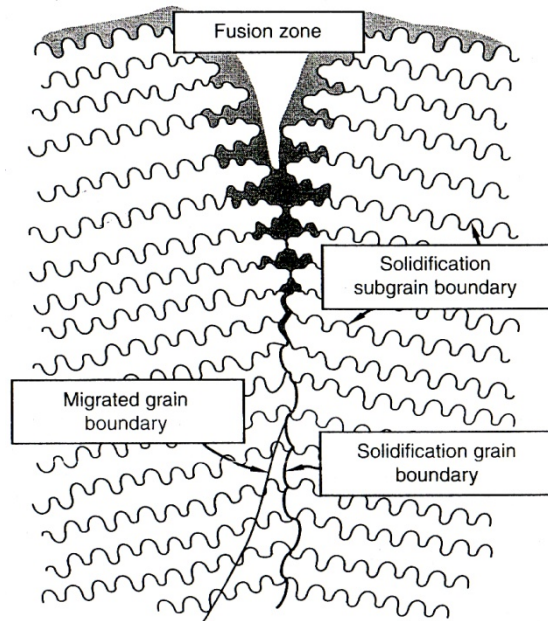


Figure 4.10 The different types of boundaries observed in austenitic stainless steels.

The **Solidification Subgrain Boundaries (SSGB)** are actually the boundaries separating adjacent subgrains that are normally present as cells or dendrites. They are crystallographically characterized as *low-angle boundaries*, because of their low angular misorientation resulting from the fact that subgrain growth during solidification occurs along preferred crystallographic directions.

Solidification Grain Boundaries (SGB) are the result of the intersection of packets or groups of subgrains. As such, they form through competitive growth that occurs during solidification along the trailing edge of the weld pool. They are characterized as *high-angle boundaries*, because of their high angular misorientation. This misorientation results in the formation of a dislocation network along the SGB. High concentrations of impurity elements at the SGB's may form low-melting liquid films along the SGB's that can promote **weld solidification cracking**, a defect that will be discussed later on.

SGB's that form at the end of solidification have both a crystallographic as well as a compositional component and in some situations, it is possible for the crystallographic component to move away from the compositional component. This results in the formation of a new boundary that carries with it the high-angle misorientation of the parent SGB and is called **migrated grain boundary (MGB)**. This migration is even easier

during reheating, such as during multipass welding. It should be noted, however, that MGB's are most prevalent in fully austenitic weld metals. When the weld metal undergoes AF solidification, ferrite forms along the SSGB's and SGB's, thus "pinning" the crystallographic component of the SGB and preventing it from migrating away from the parent SGB.

It should be noted, however, that MGB's, SGB's and SSGB's also form during solidification that occurs in the FA and F modes, but they are typically not apparent in the microstructure.

Although austenitic alloys are considered to be very weldable, they are subject to a number of weldability problems if proper precautions are not taken. One such problem is **weld solidification cracking**, that is considered to be one of the most formidable problems with austenitic stainless steels. Weld metals that solidify in the A mode are most susceptible to weld solidification cracking, while those that solidify in the FA mode tend to be more resistant, since the presence of a two-phase austenite + ferrite mixture along the SGB's at the end of solidification resists wetting by liquid films and presents a tortuous boundary, along which cracks are difficult to propagate. High impurity levels, especially sulphur and phosphorous, tend to increase the susceptibility in alloys that solidify in the A or AF mode. Weld restraint conditions and weld shape also influence cracking susceptibility. Other factors that promote solidification cracking are high heat inputs, excessive travel speeds, concave bead shape and underfilled craters at weld stops. The following figure clearly shows that weld solidification cracking is a strong function of composition:

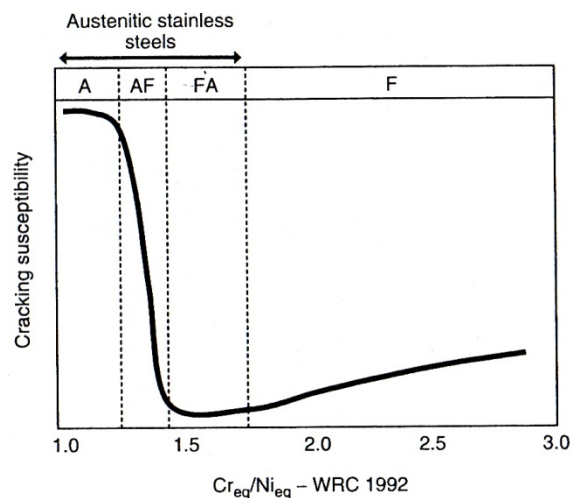


Figure 4.11 Cracking susceptibility of austenitic stainless steels, as a function of their composition.

As seen, compositions resulting in primary austenite solidification (A and AF) are most susceptible to cracking, while the FA mode offers the greatest resistance to solidification cracking. F mode is more susceptible to cracking than FA, but still superior to A and AF. Thus, controlling the composition of base and filler materials is a very effective way to prevent weld solidification cracking.

Another problem commonly encountered in austenitic stainless steel weld metals is **HAZ liquation cracking (hot cracking)**. It occurs due to the formation of liquid films along grain boundaries in the partially melted zone adjacent to the fusion boundary.

Sulphur has been found to be extremely harmful with respect to HAZ liquation cracking, but phosphorous can be very harmful too. HAZ liquation cracking can be controlled by controlling the composition of the base metal and filler metal so as to obtain a microstructure with a small amount of ferrite in the austenite matrix. The ferrite provides ferrite-austenite boundaries that are able to control the sulfur and phosphorous compounds so they do not permit hot cracking. In alloys with fully austenitic HAZ, the risk of liquation cracking can also be reduced by restricting impurity levels and grain size, as well as lowering the heat input. However, restricting impurity levels significantly increases the cost of manufacturing the steel.

The same problem can be encountered in weld metal, only this time it is referred to as **weld metal liquation cracking**. It occurs during multipass welds along SGB's or MGB's. Once again, fully austenitic welds are most susceptible to the problem, while weld metals with sufficient ferrite appear to be resistant to weld metal liquation cracking. It should also be noted that these defects are also known as *microfissures*, since they tend to be very small and usually buried within the weld deposit. Being generally short and tight, these cracks are extremely difficult to detect and can sometimes be located by grounding the top surface of the weld and performing dye penetrant inspection. Even then they appear as spots rather than cracks. To avoid weld metal liquation cracking one should control the deposit composition in such a way that some ferrite is present in the deposit. When dealing with fully austenitic deposits, however, control of impurity levels and minimizing weld heat input can reduce or eliminate this form of cracking. Some filler metals with increased Mn have also been developed to minimize cracking in fully austenitic weld metals. It should be noted though that when weld deposits with higher ferrite are reheated, the reduction in HAZ ferrite will not be sufficient to promote cracking.

In many alloys with an austenitic microstructure (FCC), including of course austenitic stainless steels, **ductility-dip cracking (DDC)** often occurs. This type of cracking is associated with a precipitous drop in elevated temperature ductility that is encountered above approximately one-half the melting temperature of the material, as shown on the following figure:

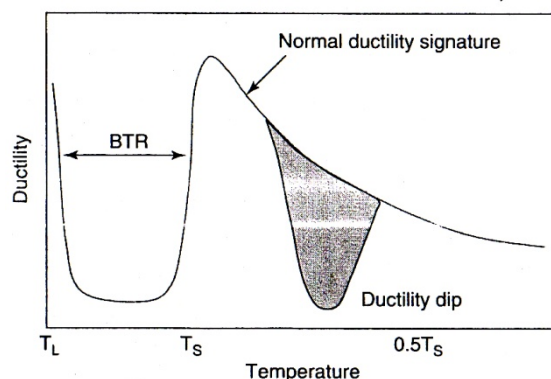


Figure 4.12 Ductility-dip cracking encountered in austenitic stainless steels.

Note that this ductility dip is separate and distinct from the brittle temperature range (BTR) in which solidification and liquation cracking occur. It is also possible that liquation cracks and DDC may actually connect to form a single crack, although by two different mechanisms. DDC in weld metals occurs along MGB's and has been observed

in both the weld metal and the HAZ of austenitic stainless steels. It is normally associated with large grain size and high restraint conditions, such as those generated in thick-section weldments. A way to reduce susceptibility to DDC is using methods that increase boundary tortuosity effectively pinning the grain boundary, e.g. AF or FA solidification.

Reheat cracking (or stress-relief cracking) is fairly unusual in standard grade austenitic stainless steels, but it may occur in alloys that form MC-type carbides during the stress-relief thermal cycle. During welding, alloy carbides dissolve in the high-temperature HAZ adjacent to the fusion boundary and the carbides precipitate in the grain interiors, upon reheating of the weldment, strengthening these regions in relation to the grain boundaries. If significant stress relaxation occurs in this same temperature range, failure can occur preferentially along the grain boundaries. This phenomenon has also been observed in both the HAZ and the weld metal. Reheat cracking normally exhibits a “C-curve” temperature-time relationship, just like the one showing below:

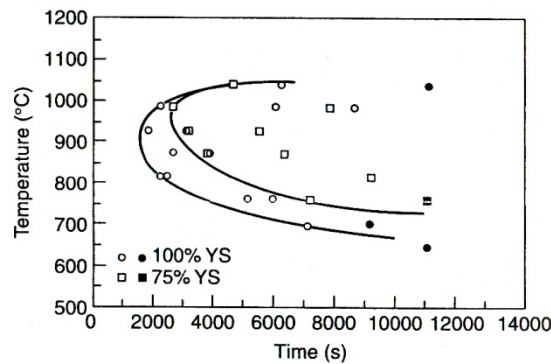


Figure 4.13 The “C-curve” temperature-time relationship of reheat cracking.

Copper contamination cracking is another well known phenomenon in austenitic stainless steels. It occurs by a liquid metal embrittlement mechanism, whereby molten copper penetrates austenite grain boundaries, as shown on the following figure:

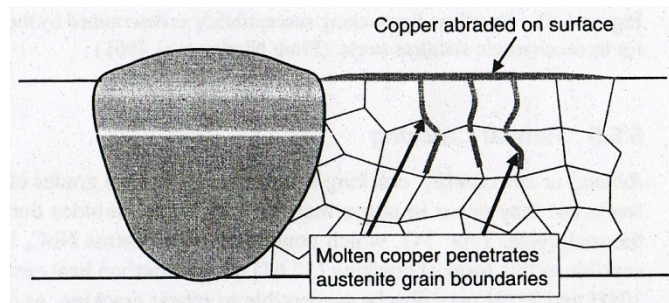


Figure 4.14 The mechanics of copper contamination cracking.

Since molten copper is required for this phenomenon to take effect, it can only occur above the melting point of copper, approximately 1083°C (1981°F). Welding-related failures due to CCC almost always result from the abrasion of copper on the parts to be welded. Copper added to the material as an alloying element does not promote CCC. This type of cracking is usually observed within a small distance from the

fusion boundary, since copper is most effective at wetting the grain boundaries when it is at temperatures above 1100°C (2010°F). However, eliminating the source of CCC is much more challenging than identifying its existence. Most times plating any copper fixtures with chromium, nickel, or other materials, thus effectively isolating the copper parts, may reduce the material's susceptibility to CCC.

Another contamination-related cracking is ***zinc contamination cracking (ZCC)***. It is very similar to CCC, but the means by which zinc infiltrates the grain boundaries is different. Zinc melts at a temperature of 419.5°C (787°F), which is much lower than copper. ZCC occurs in welding galvanized steel to austenitic stainless steel, regardless of the filler metal used, but the result stays the same; cracking along grain boundaries in the HAZ of austenitic stainless steel, a few millimeters from the fusion boundary. It is best prevented by removing the galvanizing from the mild steel in the joint area by acid dissolution before welding.

Lastly, there is also ***helium-induced cracking*** that is associated with helium bubble formation along the base metal grain boundaries. The helium bubble is a by-product of neutron irradiation. Helium has very low solubility in steel and thus tends to form small bubbles along grain boundaries and other defect sites. In the HAZ, these bubbles grow rapidly by diffusion of helium along the grain boundary. If sufficient restraint is present, cracking will occur along this boundary, in very close proximity of the fusion boundary. Unfortunately, this type of cracking is extremely difficult to avoid since helium is nearly impossible to be removed from the steel's defect sites. Some success toward that end, however, was achieved by first applying low heat input weld overlays and then making repair weld attachments to the overlay material, effectively a form of "buttering" the substrate.

4.3.3. Corrosion and Erosion Resistance^{[2] [25] [30] [31] [32] [43]}

Austenitic stainless steels are selected because of their good corrosion resistance. However, certain precautions are required whenever they are exposed to certain environments. At room temperature and in a general atmospheric environment their corrosion is practically nil. In freshwater marine environments, general corrosion rates are also quite low. At elevated temperature, however, general corrosion rates increase and some degradation and material loss will occur over time. Being a process that raises the metal's temperatures, welding can produce metallurgical modifications that tend to increase the alloy's susceptibility to corrosion attack.

There are many types of corrosion that could attack an austenitic alloy at any moment, but the most compromising for the engineering usefulness of the weld structure is ***intergranular corrosion (IGC)***, also called ***intergranular attack (IGA)***. This type of corrosion appears on the surface of the weld facing the corrosive environment and presents itself as a linear area of attack that parallels the fusion boundary. These are also called *wagon tracks* and a schematic representation is on the following figure:

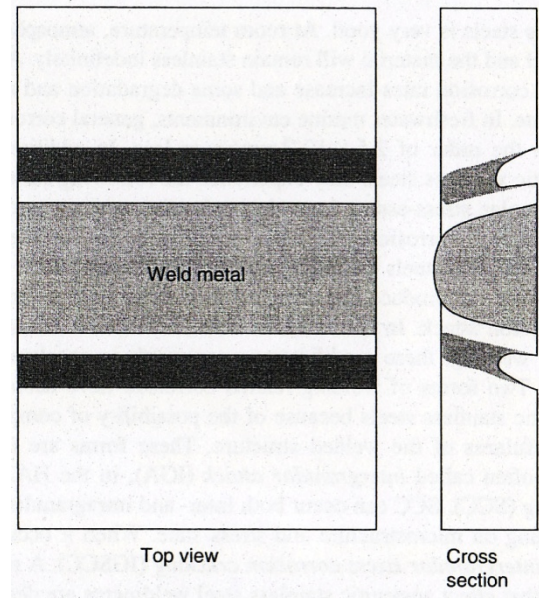


Figure 4.15 “Wagon tracks” created by intergranular corrosion.

As one can observe, there is severe attack along parallel sensitized bands in the HAZ, at some distance from the fusion boundary. This is due to the fact that the attack is actually the result of chromium carbide precipitation; carbide precipitation occurs in the temperature range of 600 to 850°C (1110 to 1560°F). Above this temperature carbides go back into solution and thus the region around the fusion boundary is relatively free of carbides. The end result, however, is a chromium-depleted zone on both sides of the HAZ that is sensitive to corrosive attack. At this point it should be noted that, except in fully austenitic stainless steels, sensitization is largely a HAZ problem and not a weld metal problem, since the ferrite that is normally found in the weld metal is richer in Cr than the austenite and Cr diffuses much more rapidly in ferrite than in austenite, which helps overcome any Cr depletion. The stabilized grades of austenitic stainless steel containing Nb and/or Ti (such as 347 and 321) exhibit very good resistance to this type of corrosion, since Nb and Ti have the ability to tie up carbon in the form of stable MC-type carbides, thus minimizing the formation (and subsequent precipitation) of chromium carbides at the grain boundaries. Carbon content is the other factor that has a profound influence on susceptibility to IGC in austenitic stainless steels, since at low carbon content high-temperature precipitation requires at least one hour to even start, while this time is severely decreased for higher carbon content. As a result, the use of L-grade (low carbon) alloys minimizes the risk of IGC. Another factor that would accelerate the precipitation reaction is the presence of residual stresses. Care must be taken during PWHT, however, not to sensitize the entire structure, since stress relief temperature range for most austenitic stainless steels overlaps the carbide precipitation temperature range.

Intergranular attack can also occur in the stabilized grades, but it normally occurs in a very narrow region just adjacent to the fusion boundary and is called **knifeline attack**. This type of attack occurs when the stabilized carbides (NbC and TiC) dissolve at elevated temperatures in the region adjacent to the fusion zone. Upon cooling, however, Cr-rich carbides form faster than the aforementioned stable carbides, resulting in a narrow sensitized region as shown on the figure below:

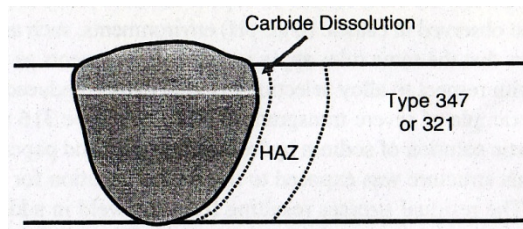


Figure 4.16 The region where knifeline attack can occur.

Another form of welding-related corrosion that presents a great deal of interest, since it could potentially prove to be very compromising for a weld structure, is **stress corrosion cracking (SCC)** to which austenitic stainless steels are inherently susceptible, particularly in Cl-bearing environments such as seawater. SCC, in general, is the unexpected sudden failure of normally ductile metals subjected to a tensile stress in a corrosive environment, especially at elevated temperature. Resistance to SCC increases with the increased content of above all nickel and molybdenum. However, it is extremely noteworthy that SCC occurs for nickel content at the range of 8 to 12wt%. That means that SCC can practically be avoided by selecting alloys with either lower (<5wt%) or higher (>20wt%) nickel content. In the former case, the use of Ferritic or Duplex stainless steels is advised. In the latter case, Superaustenitics or Ni-base alloys can be used. Regardless the case, austenitic stainless steels have a rather reduced resistance to SCC. SCC has also been observed in caustic (high-PH) environments and this type of corrosion can only be avoided by proper alloy selection. The resistance to alkaline SCC is more dependent on the nickel content of the material and in this respect *high performance austenitic steels* are superior to conventional stainless steels. Nickel-based alloys, however, are to be preferred in the most demanding conditions. Welding may exacerbate SCC in alloy systems that are otherwise resistant to this type of corrosion, because it changes their microstructures and creates residual stresses. Postweld stress relief can sometimes be used to reduce these stresses and minimize susceptibility to SCC, but extreme caution should be exercised to avoid sensitization, as noted above.

Pitting and **crevice corrosion** are corrosion attacks that are considered related phenomena. Halide element ions, most commonly Cl^- , are very often present in aqueous solutions and are definitely present in saltwater environments. A penetration of the passive film of stainless steels allows the underlying metal to become active. The penetration site may be an inclusion exposed to the corrosive media. A crevice may be present at the edge of a gasket or sealing ring and can also serve as a pit initiation site. Chromium, molybdenum, tungsten, and nitrogen are alloy elements that improve resistance to pitting and crevice corrosion. This is often illustrated using the *pitting resistance equivalent (PRE)* for the material, which can be calculated using the following formula:

$$\text{PRE} = \% \text{Cr} + 3.3 \cdot \% \text{Mo} + 16 \cdot \% \text{N}$$

The PRE value can be used for rough comparisons of different materials. A much more reliable means, however, is to rank the steel according to the *critical pitting temperature (CPT)* of the material. The following table contains indicative values of CPT for various austenitic stainless steel alloys:

Steel grade			Chemical composition, %						CPT °C
Avesta	SS	ASTM	C	Cr	Ni	Mo	N	Other	
17-11-2	2347	316	0.048	17.00	10.73	2.30	0.037		19
			0.037	16.86	10.72	2.17	0.036		12
			0.039	17.02	10.64	2.35	0.037		18
17-11-2Ti	2350	316Ti	0.033	16.77	11.24	2.05	0.011	Ti	9
			0.049	16.89	10.78	2.09	0.013	Ti	8
			0.040	16.76	10.76	2.07	0.014	Ti	10
			0.039	16.75	10.85	2.06	0.011	Ti	5
			0.057	16.77	10.72	2.15	0.013	Ti	10
17-11-2L	2348	316L	0.014	17.53	10.90	2.15	0.069		16
			0.024	17.39	10.46	2.05	0.062		15
			0.023	16.80	11.19	2.16	0.063		13
			0.017	16.77	11.24	2.09	0.047		10
17-11-2LN	-	316LN	0.018	17.22	11.03	2.41	0.174		24
			0.018	17.19	10.62	2.06	0.136		26
17-12-2.5	2343	316L	0.017	16.67	11.65	2.73	0.058		13
			0.048	16.74	10.82	2.57	0.037		18
			0.045	17.43	10.79	2.67	0.068		32
			0.040	17.54	10.59	2.57	0.070		28
			0.045	16.70	10.75	2.54	0.042		12
			0.045	17.27	10.61	2.68	0.066		24
			0.056	17.56	11.23	2.58	0.068		28
			0.046	16.89	10.75	2.74	0.052		26
			0.047	17.47	11.10	2.55	0.068		28
17-12-2.5L	2353	316L	0.018	17.30	12.63	2.59	0.052		22
			0.023	17.27	12.48	2.46	0.070		22
			0.020	17.29	12.55	2.72	0.066		21
			0.019	17.25	12.61	2.56	0.054		19
			0.030	17.37	12.75	2.58	0.050		19
17-12-2.5LN	2375	316LN	0.017	17.22	12.61	2.60	0.058		18
			0.020	17.59	12.28	2.61	0.142		34
			0.016	17.24	12.21	2.55	0.167		30
18-13-3L	2367	317L	0.022	17.59	12.28	2.61	0.144		35
			0.023	18.33	13.21	3.18	0.089		43
			0.015	18.09	13.19	3.10	0.065		32

Figure 4.17 Indicative values of critical pitting temperature (CPT) for various austenitic stainless steel alloys.

It should also be noted here, since our main concern in this thesis is about shipbuilding applications of stainless steels, that natural seawater contains living organisms which tend to form a biofilm on stainless steel. This film increases the corrosion potential of the steel and thus, also the risk of pitting and crevice corrosion. The activity of the biofilm is temperature-related, but since the different organisms are adapted to the natural temperature of the water, their activity varies between different seas around the world. This means that in cold seas the natural water is most aggressive at 25-30°C while the corresponding value in tropical seas is just above 30°C. The biological activity ceases at higher temperatures. In many seawater systems the water is chlorinated with either chlorine or hypochlorite solutions to reduce the risk of fouling. Both chlorine and hypochlorite are strongly oxidizing agents and they cause the corrosion potential of the steel surface to exceed what is normal in non-chlorinated seawater, which in turn means increased risk of corrosion. In chlorinated seawater the aggressiveness increases as the temperature rises.

When dealing with austenitic stainless steels, one should also consider **sulphur attack**. Various sulphur compounds are often present in gases and, as a rule, they have a very detrimental effect on the service life of the exposed components. Sulphides can nucleate and grow due to kinetic effects even under conditions where only oxides would

form from a thermodynamic point of view. In existing oxide layers, attacks can occur in pores and cracks. It is therefore essential that the material is able to form a thin, tough, and adherent oxide layer. This requires high chromium content and preferably also additions of silicon, aluminum, and/or reactive elements. Under so-called reducing conditions, the oxygen activity of the gas can still be sufficiently high to enable the formation of a protective oxide layer, provided that the chromium content of the material is sufficiently high (>25%). If this is not the case, low-melting-point nickel sulphides can be formed instead. Under such circumstances, a nickel-free (or low nickel) material should be selected.

There are also cases, in certain aqueous environments, where aerobic bacteria literally attack the metal. This corrosive attack is called ***microbiologically induced corrosion (MIC)***. MIC manifests itself as pitting, whereby metal surface exhibits a small hall or pit, but it can be distinguished from pitting or crevice corrosion by the presence of “tubercles” of biological residue and corrosion products over the pit. This form of corrosion has been observed in both fresh water and seawater and requires oxygenated water to support the metal dissolution reaction. Austenitic stainless steels seem to be very susceptible and the presence of a two-phase austenite + ferrite microstructure, such as that present in many weld metals, seems to increase susceptibility.

Austenitic weld metals that contain a small amount of ferrite have corrosion resistance equivalent to their corresponding ferrite-free base metals in most corrosive environments. However, there are certain environments in the manufacture of urea, or organic acids such as pteraphthalic acid, which selectively attack ferrite in nominally austenitic stainless steel welds and overlays. This phenomenon is often referred to as ***selective ferrite attack*** and manifests itself in cases where ferrite is present in any appreciable amount (mainly greater than 5FN), at which point corrosion will proceed along ferrite networks and can effectively destroy the weld metal. The only way to avoid this type of attack is using virtually ferrite-free alloys and filler metals, such as 316L and 18Cr-15Ni-3Mo L and 18Cr-16Ni-5MoN L respectively.

4.4 SHIPBUILDING APPLICATIONS^[2]

It is considered really important to point out that common austenitic stainless steels are considered a poor choice for seawater environments or other chloride-containing media, or in highly caustic environments. These grades tend to be very susceptible to stress corrosion cracking, when encountered in such environments, since the phenomenon afflicts the base metal, HAZ, and weld metal in these alloys.

However, austenitic stainless steel can be used for cladding operations in crude oil carriers, to help protect against corrosion from H₂S atmospheres.

FERRITIC STAINLESS STEELS: Fe-Cr-(Mo) alloys

5.1 GENERAL ^{[1] [2] [16] [25] [26] [27]}

Historically, ferritic stainless steels had been used in the greatest tonnage in applications that did not require welding, such as automotive trim and other decorative/architectural applications. Since the early 1980's the use of low- and medium-Cr grades for automotive exhaust systems has increased dramatically. The fact that ferritic alloys started being used for applications that required welding intensified the research efforts towards their weldability. Various high-Cr grades have been developed ever since for use in demanding environments, possessing superior corrosion resistance in relation to austenitic and martensitic grades. However, these alloys are relatively expensive and difficult to fabricate.

In general, ferritic stainless steels are characterized as such because their predominant metallurgical phase is ferrite. These alloys are, in principle, ferritic at all temperatures and this is achieved by a low content of austenite-formers, mainly nickel, and a high content of ferrite-formers, mainly chromium. Their chemical composition ($0.02 \leq C \leq 0.12$, $6.00 \leq Cr \leq 30.00$, $0.10 \leq Mo \leq 4.50$) gives them good resistance to stress corrosion cracking, pitting corrosion, and crevice corrosion (especially in chloride environments). They have slightly higher yield strength than austenitic steels, but they have less elongation at fracture and much lower strain hardening. They are used in a variety of applications where corrosion resistance, rather than mechanical properties, is the primary concern.

Ferritic grades are generally limited to service temperatures below 400°C (750°F), due to the risk of embrittling phases formation, such as 475°C (885°F) embrittlement,

that is the bane of high-Cr grades. Ferritic stainless steels are ferromagnetic and cannot be hardened or strengthened through any heat treatment. They can be, however, cold worked and softened by annealing.

The older type ferritic alloys were mainly used for household utensils and other purposes where corrosion conditions were not particularly demanding. Today's low-Cr grades (10.5 to 12.5wt%) are used for applications such as automotive exhaust systems and hot water tanks, where resistance to general corrosion is superior to carbon steels. Medium- and high-Cr grades are used in more aggressive environments, while Superferritic grades are used in the chemical processing and pulp and paper industries, where resistance to corrosion in severe oxidizing media is required. High-Cr grades are also used in high-efficiency furnaces.

5.2 COMPOSITION AND MECHANICAL PROPERTIES^{[2] [25] [27] [29] [30]}

As stated above, ferritic grades have been developed to provide a group of stainless steel to resist corrosion and oxidation, while still being resistant to the bane of all stainless steels; stress corrosion cracking. As a group they are more corrosive resistant than the Martensitic grades, but still not as resistant as the Austenitic grades. Like Martensitic stainless steels, they are straight chromium steels with no nickel and chromium content that ranges mainly from 10.5wt% to 30wt%. Their carbon content does not exceed 0.20wt% and sometimes they also contain ferrite promoters like Al, Nb (Cb), Ti and Mo.

Ferritic grades have evolved through the years in three generations, depending on their composition ranges. First-generation alloys were mainly the medium-chromium types with relatively high-carbon content. They are not 100% ferritic in composition, since during solidification and cooling or when heated to elevated temperature they form some austenite, which then transforms to martensite upon cooling to room temperature.

Second-generation ferritic alloys have been developed to minimize the exact same formation of martensite in the ferrite structure, so that weldability can be improved. These grades have lower-carbon contents and often contain a stabilizing element, such as Nb or Ti, which ties up carbon and nitrogen, thereby preventing the formation of unstable carbides and promoting ferrite stability.

Third-generation ferritic alloys have high chromium content, but low interstitial (carbon and nitrogen) levels, and low impurity levels. The high-purity grades of ferritic stainless steels have superior corrosion resistance with moderate toughness and ductility, and when welding them one should use extreme caution to avoid pickup of undesirable elements (mainly nitrogen and oxygen) and to minimize grain growth. These grades are often developed for specific applications.

The following figure illustrates how alloying elements can determine the types of the ferritic group:

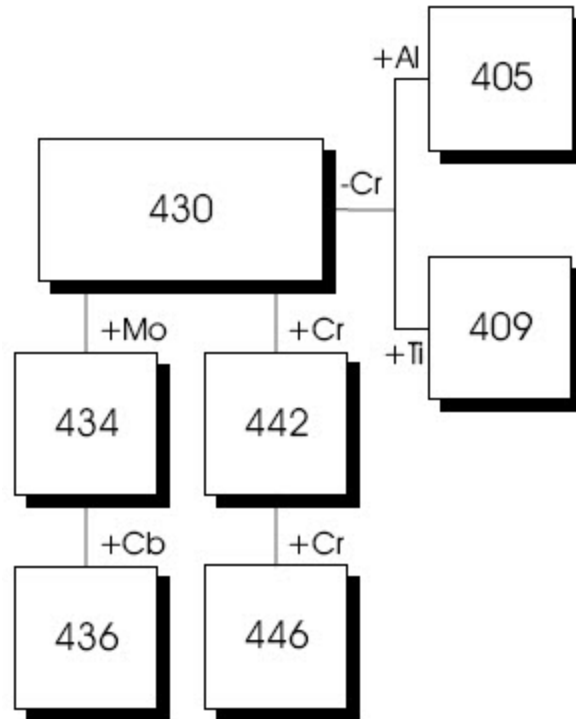


Figure 5.1 The Ferritic Group.

Type 430 is the basic ferritic grade, with a little less corrosion resistance than Type 304 Austenitic stainless steel. It combines high resistance to corrosive environments such as nitric acid, sulphur gases, and many organic and food acids. Type 405 has lower chromium content and added aluminum to prevent hardening when cooled from high temperatures. Its typical applications include heat exchangers. The grade having the lowest chromium content of all stainless steels is Type 409, but it is also the least expensive of all steels. It was originally designed for muffler stock and was also used for exterior parts in non-critical corrosive environments. Type 434 has molybdenum added for improved corrosion resistance. Type 436 has columbium added for even better corrosion and heat resistance. Some of its typical applications include deep-drawn parts. Type 442 has increased chromium to improve scaling resistance and is mainly used for furnace and heater parts. Finally, Type 446 contains even more chromium to further improve its corrosion and scaling resistance at high temperatures. It is especially good for oxidation resistance in sulphuric atmospheres.

The following table lists detailed compositions of ferritic stainless steels:

Type	UNS Number	Composition - Percent *							Other
		C	Mn	Si	Cr	Ni	P	S	
405	S40500	0.08	1.00	1.00	11.5-14.5		0.04	0.03	0.10-0.30 Al
409	S40900	0.08	1.00	1.00	10.5-11.75		0.045	0.045	6 x %C min. Ti
429	S42900	0.12	1.00	1.00	14.0-16.0		0.04	0.03	
430	S43000	0.12	1.00	1.00	16.0-18.0		0.04	0.03	
430F**	S43020	0.12	1.25	1.00	16.0-18.0		0.06	0.15 min.	0.06 Mo
430FSe**	S43023	0.12	1.25	1.00	16.0-18.0		0.06	0.06	0.15 min. Se
430Ti	S43036	0.10	1.00	1.00	16.0-19.5	0.75	0.04	0.03	5 x %C - Ti min.
434	S43400	0.12	1.00	1.00	16.0-18.0		0.04	0.03	0.75-1.25 Mo
436	S43600	0.12	1.00	1.00	16.0-18.0		0.04	0.03	0.75-1.25 Mo; 5 x %C min. Nb(Cb) + Ta
442	S44200	0.20	1.00	1.00	18.0-23.0		0.04	0.03	
444	S44400	0.025	1.00	1.00	17.5-19.5	1.00	0.04	0.03	1.75-2.5 Mo, 0.035 N 0.2 + 4 (%C + %N); (Ti + Nb(Cb))
446	S44600	0.20	1.50	1.00	23.0-27.0		0.04	0.03	0.25 N
18-2FM**	S18200	0.08	2.50	1.00	17.5-19.5		0.04	0.15 min.	
18SR		0.04	0.3	1.00	18.0				2.0 Al; 0.4 Ti
26-1 (E-Brite)	S44625	0.01	0.40	0.40	25.0-27.5	0.50	0.02	0.02	0.75-1.5 Mo; 0.015N; 0.2 Cu; 0.5 (Ni+Cu)
26-1Ti	S44626	0.06	0.75	0.75	25.0-27.0	0.5	0.04	0.02	0.75-1.5 Mo; 0.04 N; 0.2 Cu; 0.2-1.0 Ti
29-4	S44700	0.01	0.30	0.20	28.0-30.0	0.15	0.025	0.02	3.5-4.2 Mo
29-4-2	S44800	0.01	0.30	0.20	28.0-30.0	2.0-2.5	0.025	0.02	3.5-4.2 Mo
Monit	S44635	0.25	1.00	0.75	24.5-26.0	3.5-4.5	0.04	0.03	3.5-4.5 Mo; 0.3-0.6 (Ti + Nb(Cb))
Sea-cure/ Sc-1	S44660	0.025	1.00	0.75	25.0-27.0	1.5-3.5	0.04	0.03	2.5-3.5 Mo; 0.2 + 4 (%C + %N) (Ti + Nb(Cb))

Figure 5.2 Ferritic stainless steels composition table.

It should be noted that many of the alloys in this group have a 4XX designation, similar to that of the Martensitic stainless steels, even though their microstructure and properties are different.

Mechanical properties, corrosion resistance and weldability, as said before, are greatly dependent on the steel's microstructure. The presence of other elements, either as additions or as impurities, can have a significant effect on the microstructure of ferritic stainless steels. Nitrogen, normally present as an impurity rather than an intentional addition, has an effect similar to that of carbon in promoting austenite formation. Various levels of carbon and nitrogen have a huge effect on the expansion of the gamma loop in simple Fe-Cr alloys. As a result, either reduction of nitrogen and carbon to extremely low levels, or addition of alloying elements that promote ferrite formation is necessary to maintain a primarily ferritic microstructure in low- and medium-chromium steels. Other ferrite-promoting elements, other than chromium, that are commonly added to ferritic stainless steels include silicon, titanium, niobium, molybdenum and aluminum. Titanium and niobium are very useful in small concentrations due to their affinity for both carbon and nitrogen, while aluminum is effective in combining with nitrogen. Another effect of aluminum addition is improved oxidation resistance, particularly at elevated temperature. Silicon is normally added as a deoxidizer and to provide oxide-scaling resistance. Molybdenum is added to some alloys to improve pitting corrosion resistance. Austenite-promoting elements include manganese, nickel and copper in addition to carbon and nitrogen. Manganese is traditionally added to control sulphur and therefore to improve the steel's castability

and subsequent hot-working characteristics. Nickel and copper are not intentionally added to ferritic stainless steels, although small amounts of nickel may be effective in improving notch toughness.

Another point that should be discussed is that under normal thermo-mechanical processing conditions, any austenite that forms at elevated temperature will generally transform to martensite upon cooling to room temperature and the existence of martensite in ferritic stainless steels has been shown to have both beneficial and detrimental effects. For example, the presence of martensite in appreciable quantities has been purported to promote hydrogen-induced cracking and it has also been cited as a source of embrittlement due to its fracture and deformation characteristics relative to ferrite. However, the elevated temperature microstructure does not allow martensite to reach high enough levels to cause any loss of ductility and toughness. As such, some martensite can be accommodated without a significant penalty with respect to mechanical properties and, as a matter of fact, martensite formation in the weld metal and HAZ of low- and medium-chromium ferritic stainless steels is quite common, even though its presence in low-chromium ferritic alloys has been related to a loss of corrosion resistance.

Before delving any further in the mechanical properties of ferritic stainless steels, however, it should be pointed out that there are three embrittlement phenomena that influence the mechanical properties of these alloys: **475°C (885°F) embrittlement** and **sigma phase precipitation**, that are considered to be **intermediate-temperature embrittlement (ITE)** phenomena, and **high temperature embrittlement (HTE)**.

Starting with the first type of embrittlement, Fe-Cr alloys containing from 15 to 70wt% chromium can be severely embrittled when heated into the temperature range of 425 to 550°C (800 to 1020°F). The reason for this is still highly controversial, but the predominant theory associates the onset with the formation of a coherent precipitate at temperatures below 550°C (1020°F), due to the presence of a miscibility gap in the Fe-Cr equilibrium phase diagram. The rate and degree of this type of embrittlement is a function of chromium content, with high-chromium alloys embrittling in much shorter times and at slightly higher temperatures. The lowest Cr ferritic stainless steels, such as Types 405 and 409, seem to be immune to this type of embrittlement, while aging times of at least 100 hours are required to cause embrittlement in low- and medium-chromium alloys. High-chromium alloys may exhibit loss of ductility and toughness after much shorter times. Alloying additions such as molybdenum, niobium and titanium tend to accelerate the onset of 475°C embrittlement, as does cold work. This type of embrittlement also results in severe reduction in corrosion resistance, probably due to selective attack of the iron-rich ferrite. 475°C embrittlement can be eliminated and mechanical and corrosion properties restored to the unaged condition by heating the alloy into the range of 550 to 600°C (1020 to 1110°F) for a short time.

Excessive time in this temperature range, however, will actually produce sigma phase embrittlement. As with 475°C embrittlement, alloys with higher Cr concentrations are more susceptible to sigma formation and the rate of formation is more rapid. In alloys containing less than 20wt% chromium, sigma phase does not form readily, often requiring hundreds of hours of exposure in the critical temperature range. In high-Cr alloys, sigma phase formation is more rapid, requiring only a few hours of exposure in the sigma formation temperature range. All this is perfectly illustrated in the following figure:

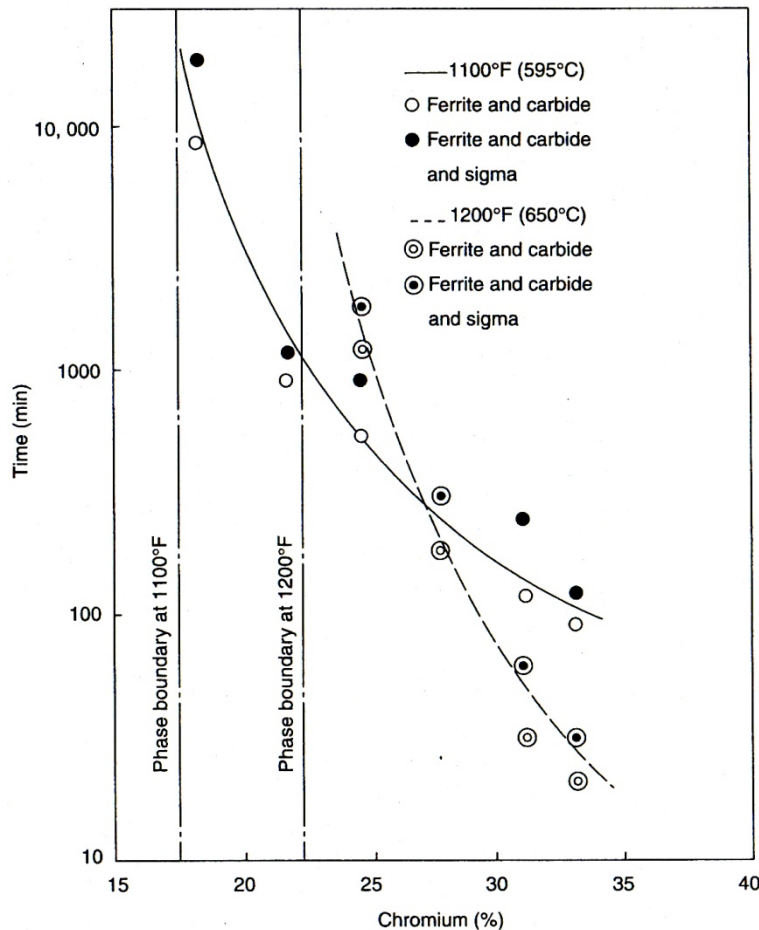


Figure 5.3 Sigma phase formation as a function of temperature, time, and chromium content.

Alloying additions such as molybdenum, nickel, silicon and manganese shift the sigma formation range to higher temperatures, lower chromium concentrations and shorter times. As with other precipitation phenomena, cold work accelerates the formation of sigma phase. The deleterious effects of sigma precipitation can be eliminated by heating for short time to temperatures above 800°C (1470°F).

Sigma phase precipitation and 475°C embrittlement are not normally a problem when welding ferritic stainless steels, since embrittlement due to these phenomena is normally associated with long exposure times at intermediate temperatures, hence the term *intermediate-temperature embrittlement*. ITE is actually more an issue of engineering application and limiting service exposure temperatures to below 400°C (750°F) practically avoids the problem of ITE. As stated before, both these embrittlement phenomena are accelerated as chromium content of the alloy or filler metal increases, and special precautions may be required when considering PWHT of the high-chromium alloys.

High-temperature embrittlement, on the other hand, results from metallurgical changes that occur during exposure to temperatures above approximately $0.7 \cdot T_m$ (melting temperature), which actually means that it is most common during thermo-mechanical processing or welding. Exposure to these elevated temperatures can also

result in a severe loss in corrosion resistance. Susceptibility to HTE is influenced primarily by composition (particularly chromium), interstitial concentration and grain size. Low-chromium and stabilized grades are relatively insensitive to HTE. Elimination of HTE may be possible in high-interstitial alloys by heating between 730 and 790°C (1350 and 1450°F). This heat treatment probably acts to overage the precipitates and thus reduces the deleterious effect on toughness and ductility. Some caution should be exercised when performing such a heat treatment, however, since sigma phase may precipitate during extended exposure in this temperature range. In low-interstitial alloys embrittlement occurs primarily due to grain growth at high-temperature, so thermal treatments will be of little help.

Even in the absence of ITE or HTE, however, ferritic stainless steels exhibit significant **notch sensitivity**. For low-chromium alloys in the annealed condition, impact toughness does not appear to be influenced by interstitial content. For medium- and high-chromium alloys interstitial content plays a much more critical role. In the absence of metallurgical variables such as grain size and precipitation effects, composition influences notch sensitivity significantly. In medium- and high-chromium alloys, notch sensitivity can be reduced by maintaining extremely low interstitial levels, and thus most commercial alloys are formulated with extremely low carbon and nitrogen contents or contain “control” elements, such as titanium, niobium and aluminum to neutralize the interstitials. Extreme care must be taken in the fabrication of these alloys to avoid contributory effects due to HTE.

What can be summarized from what was stated above is that reduction of mechanical and/or corrosion properties or catastrophic failure of welded ferritic stainless steels is linked almost exclusively to HTE, notch sensitivity, or a combination of the two.

If more detailed cases were to be discussed about the mechanical properties of ferritic stainless steels, as one could clearly see on the following stress-strain curve, ferritic alloys (just like Duplex stainless steels) lie somewhere between the two extremes that are set by Austenitic and Martensitic stainless steels:

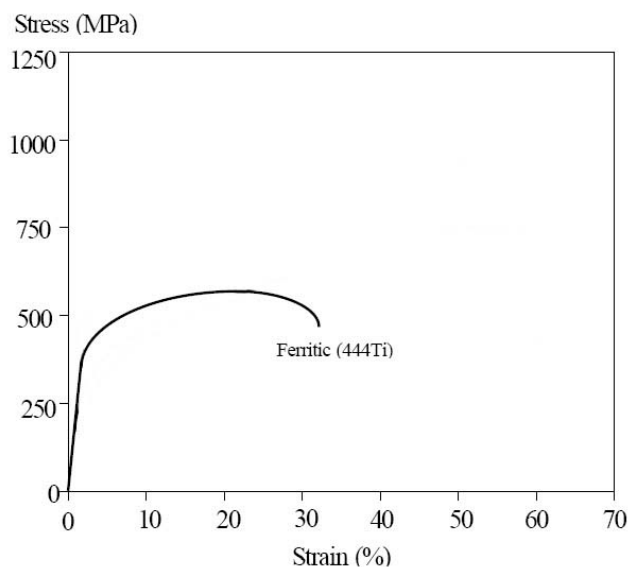


Figure 5.4 Stress-strain curve for the Type 444Ti ferritic stainless steel.

Ferritic alloys have somewhat higher yield strength than Austenitic grades and their ductility is approximately of the same order of magnitude as that of Duplex grades. Still, however, the yield strength of ferritic steels is relatively low and the work hardening is severely limited for them. Their strength can be increased with increasing carbon content, but the effect of chromium content is negligible. However, ductility decreases at high chromium levels and good ductility requires very low levels of carbon and nitrogen.

Following the general rule that toughness increases with increasing temperature, ferritic alloys, same as Martensitic and Duplex steels, are characterized by a transition in their toughness, from tough to brittle, at their transition temperature. That is shown on the following figure:

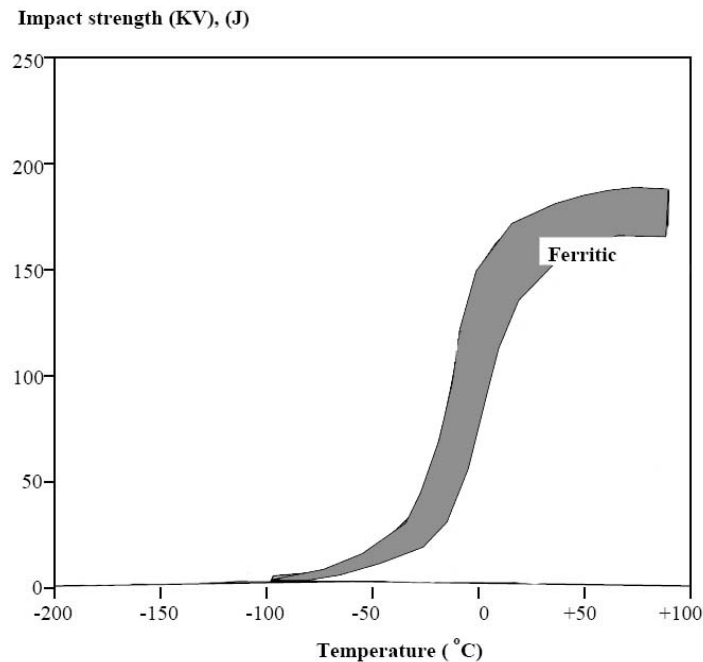


Figure 5.5 Impact toughness for ferritic stainless steels.

For ferritic alloys this transition temperature increases with increasing carbon and nitrogen content, which practically means that the steels becomes increasingly brittle at successively higher temperatures.

The fatigue strength of ferritic stainless steels is sensitive to the service environment and under both cyclic loading and corrosive conditions, corrosion fatigue, the fatigue strength will generally decrease. The relation between the fatigue limit and the tensile strength of ferritic grades can be seen on the following table:

Steel category	S_o/R_m Stress Ratio		Maximum stress
	R = -1	R = 0	
Ferritic	0.7	0.47	yield strength

The following figure, yet again, illustrates the high temperature strength of ferritic stainless steels, where the dashed line shows the elevated temperature strength of some very high alloyed and nitrogen alloyed Austenitic steels for comparison:

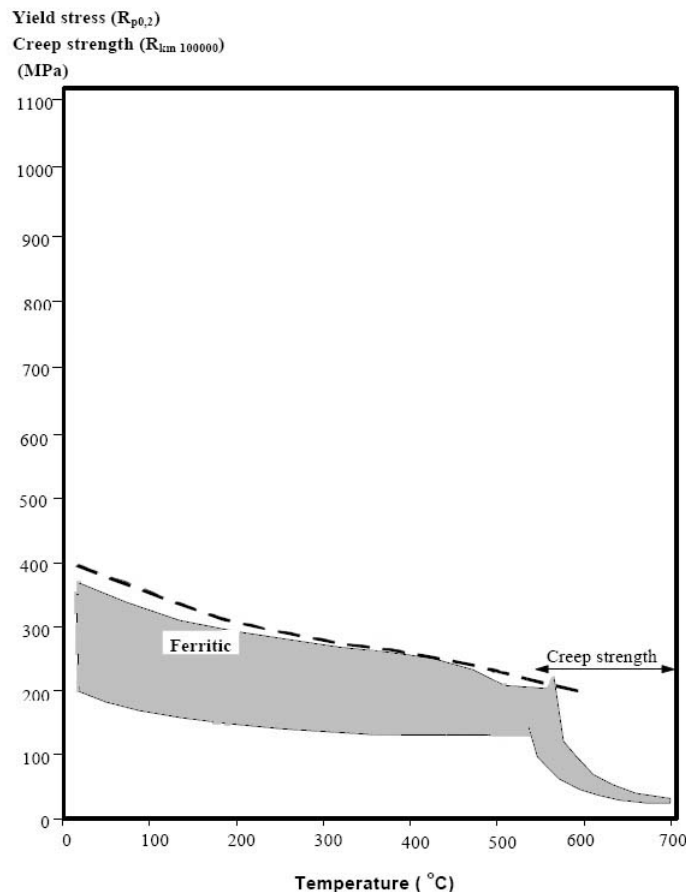


Figure 5.6 A high temperature strength diagram for ferritic stainless steels.

It is fairly obvious that ferritic stainless steels have relatively high strength up to a temperature of 500°C (935°F). The creep strength, which is usually the determining factor at temperatures above 500°C, is low. The normal upper service temperature limit is set by the risk of embrittlement at temperatures above 350°C (665°F). However, due to the good resistance of chromium steels to high temperature sulphidation and oxidation a few high-chromium grades are used in the creep range. In these cases special care is taken to ensure that the load is kept to a minimum.

The following table is derived from **ASM Metals Handbook, 8th edition, volume 1**, **ASM Metals Handbook, 9th edition, volume 3** and **ASTM standards** and synthesizes the mechanical properties of ferritic stainless steels:

Type	Condition	Tensile Strength		0.2% Yield Strength		Elong.	R.A.	Hardness
		Ksi	MPa	Ksi	MPa	%	%	Rockwell
405	Anneal	70	480	40	275	30	60	B80
409	Anneal	65	450	35	240	25		B75M
429	Anneal	71	490	45	310	30	65	B88M
430	Anneal	75	515	45	310	30	60	B82
430F	Anneal	80	550	55	380	25	60	B86
430Ti	Anneal	75	515	45	310	30	65	
434	Anneal	77	530	53	365	23		B83M
436	Anneal	77	530	53	365	23		B83M
442	Anneal	80	550	45	310	25	50	B85
444	Anneal	60	415	40	275	20		B95M
446	Anneal	80	550	50	345	23	50	B86
26-1EBrite	Anneal	65	450	40	275	22		B90M
26-1Ti	Anneal	68	470	45	310	20		B95M
29-4	Anneal	80	550	60	415	20		B98M
29-4-2	Anneal	80	550	60	415	20		B98M
18SR	Anneal	90	620	65	450	25		B90
Monit	Anneal	94	650	80	550	20		B100M
Sea-cure/SC-1	Anneal	80	550	55	380	20		B100M

Figure 5.7 The mechanical properties of ferritic stainless steels.

5.3 PHYSICAL PROPERTIES^{[1] [2] [25] [29] [30] [37]}

5.3.1. General^[37]

GRADES	Physical properties of stainless steels								
	DESIGNATIONS (▼)		DENSITY	MODULUS OF ELASTICITY	MEAN COEFFICIENT OF THERMAL EXPANSION		THERMAL CONDUCTIVITY	SPECIFIC HEAT	ELECTRICAL RESISTIVITY
	EN [N°]	AISI/ASTM	at 20°C [kg/dm³]	at 20°C [GPa]	[10⁻⁶/°C]		at 20°C [W/(m·K)]	at 20°C [J/(kg·K)]	at 20°C [Ωmm²/m]
FERRITIC	1.4003 ^(b)		7,7	220	10,8	11,6	25	430	0,60
	1.4512 ^(b)	409	7,7	220	11,0	12,0	25	460	0,60
	1.4516 ^(b)		7,7	220	10,5 ^(b)	11,5 ^(b)	30	460	0,60
	1.4000 ^(b)	410S	7,7	220	11,0	12,0	30	460	0,60
	1.4002 ^(b)	405	7,7	220	11,0	12,0	30	460	0,60
	1.4589 ^(b)		7,7	220	11,0	12,0	25	460	0,60
		429 ^(b)	7,78	200	10,3 ^(b)	(n.r.)	25,7 ^(b)	460	0,59
	1.4595 ^(b)		7,7	220	10,8	11,6	30	460	0,60
	1.4016 ^(b)	430	7,7	220	10,0	10,5	25	460	0,60
	1.4105 ^(b)	430F	7,7	220	10,5	10,5	25	460	0,70
	1.4520 ^(b)		7,7	220	10,8	11,6	20	430	0,70
	1.4511 ^(b)		7,7	220	10,0	10,5	25	460	0,60
	1.4017 ^(b)		7,7	220	10,2 ^(b)	10,8 ^(b)	30	460	0,70
	1.4113 ^(b)	434	7,7	220	10,5	10,5	25	460	0,70
	1.4513 ^(b)		7,7	220	10,5	10,5	25	460	0,70
FERRITIC	1.4510 ^(b)	439	7,7	220	10,0	10,5	25	460	0,60
		442 ^(b)	7,78	200	10,2 ^(b)	(n.r.)	21,7 ^(b)	460	0,64
	1.4521 ^(b)	444	7,7	220	10,8	11,6	23	430	0,80
	1.4526 ^(b)	436	7,7	220	11,7 ^(b)	12,1 ^(b)	30	440	0,70
	1.4509 ^(b)		7,7	220	10,0	10,5	25	460	0,60
	1.4590 ^(b)		7,7	220	(n.r.)	(n.r.)	26	460	0,60
	1.4749 ^(b)	446	7,7	200 ^(b)	10,0	11,0	17	500	0,70
	1.4713 ^(b)		7,7	(n.r.)	11,5	12,0	23	450	0,70
	1.4724 ^(b)		7,7	(n.r.)	10,5	11,5	21	500	0,75
	1.4782 ^(b)		7,7	(n.r.)	10,5	11,5	17	500	1,10
	1.4592 ^(b)		7,7	220	(n.r.)	(n.r.)	17	440	0,67

FERROMAGNETIC

Figure 5.8 The physical properties of ferritic stainless steels.

5.3.2. Weldability and Issues^{[2] [25] [30]}

Ferritic stainless steels can be welded using any common fusion and resistance welding methods. The fusion welding methods commonly used are:

- Shielded Meta Arc Welding (SMAW)
- Gas Tungsten Arc Welding TIG (GTAW)
- Gas Metal Arc Welding MIG (GMAW)
- Oxy Fuel Welding (OFW)
- Resistance welding methods also used for welding ferritic alloys are:
- Flash Welding
- High Frequency Resistance Welding
- Percussion Welding
- Projection Welding
- Resistance Seam Welding
- Resistance Spot Welding
- Upset Welding

There are a number of factors that could cause defects in a weldment and to understand them one should be aware of certain aspects of ferritic grades welding metallurgy. As far as the alloy's fusion zone is concerned, its microstructure may either be fully ferritic or consist of a mixture of ferrite and martensite, with the martensite located at the ferrite grain boundaries. Three solidification and transformation sequences are possible for ferritic stainless steels. The first one, and the simplest, referred to as **transformation path 1** from here on after, is described as follows:



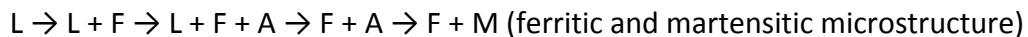
This path is dominant when the ratio of ferrite-promoting to austenite-promoting elements is high and austenite formation at elevated temperature is suppressed completely. Transformation path 1 is common in low-chromium alloys when carbon content is low, in medium-chromium alloys when Ti and Nb are added as carbide stabilizers effectively negating the austenite-promoting tendency of carbon, and in high-chromium alloys where the high-chromium content dominates. Because of the absence of elevated-temperature austenite, ferrite grain growth upon cooling from the solidification temperature range can be quite dramatic, particularly in high heat input welds.

If martensite is present in the fusion zone, however, there are two different paths the transformation could follow. **Transformation path 2** is described as follows:



Upon cooling, some austenite forms at elevated temperature along the ferrite grain boundaries. This austenite then transforms to martensite as the fusion zone cools to room temperature. This reaction can have important implications with respect to ferrite grain growth, since growth of the ferrite grains will stop once austenite starts to form along their boundaries.

The third path, **transformation path 3**, is described by the following sequence:



The solidification here begins once again as primary ferrite, but some austenite forms at the end of solidification by way of a complicated peritectic-eutectic reaction. Upon rapid cooling to room temperature, the austenite transforms once again to martensite. Transformation path 3 will generally result in a higher-volume fraction of martensite in the fusion zone microstructure relative to transformation path 2. However, the formation of austenite at the end of solidification and its presence in the solid state at elevated temperatures will restrict ferrite grain growth even more than in path 2, where a fully ferritic structure exists over some temperature range. It is within this fully ferritic high-temperature range that ferrite grain growth occurs very rapidly.

Another phenomenon observed in ferritic alloys fusion zones is their tendency to precipitate, primarily, chromium-rich carbides and chromium-rich nitrides. A similar precipitation is observed in the HAZ. Upon cooling, precipitates may form both inter- and intra-granularly, with the site dependent on cooling rate. At high cooling rates, significant intragranular precipitation is observed, while at slow cooling rates grain boundaries sites are preferred. The main reason for this precipitation is the fact that the solubility of carbon in ferrite drops dramatically as the fusion zone cools in the solid

state. In commercial low- and medium-chromium ferritic stainless steels carbide precipitation is inevitable in the absence of stabilizing elements, such as Ti and Nb, and/or elevated-temperature austenite. Many modern ferritic stainless steels contain extremely low carbon to avoid extensive carbide precipitation, but even in these steels some carbide precipitation is expected. Nitrogen behaves in a similar fashion and since most commercial first-generation ferritic alloys contain some nitrogen, precipitation of nitrogen-rich precipitates within the ferrite is also expected. Similar to carbon, the precipitation of nitrides or carbonitrides can be avoided by the addition of stabilizing elements, such as Ti and Al. If austenite is present at elevated temperature, however, precipitation will not occur due to its high solubility for carbon and nitrogen. Thus, austenite acts as a “sink” for these interstitial elements at elevated temperature and as a result when austenite is present at elevated temperature, there will often be a precipitate-free zone in the ferrite just adjacent to the martensite. Precipitation of Cr-rich precipitates, however, can also occur in the ferrite-ferrite or ferrite-martensite boundary, leading to a local reduction in chromium adjacent to the boundary and making the boundary “sensitive” to corrosion attack.

The same microstructure of ferrite and carbides is also present in the HAZ surrounding the fusion zone and it can have a huge effect on the properties of the welded metal. When this microstructure is heated to elevated temperature any carbides and other precipitates will tend to dissolve. If any martensite is present it will also tend to dissolve either by reverting to ferrite or by transforming back to austenite during reheating due to deposition of subsequent passes. The resulting elevated-temperature HAZ can either be fully ferritic or a mixture of ferrite and austenite. In the absence of precipitates and grain boundary austenite, the ferrite grains will grow quite rapidly and the HAZ of most ferritic grades exhibits relatively large ferrite grains. If austenite is stable in the microstructure at elevated temperature, it can act to inhibit ferrite grain growth by pinning down the grain boundaries.

The properties of welded ferritic stainless steels are dependent on the alloy type and its chromium concentration. As such, the most significant microstructural effect on the mechanical properties of low-chromium alloys, considering the absence of martensite, is ferrite grain growth. Thomas and Apps had the opinion that the HAZ properties of these alloys can be improved by reducing the as-welded grain size. In medium-chromium alloys mechanical behavior is also dependent on microstructure. However, a differentiation must be made between ferritic and duplex ferritic/martensitic structures, since in the former case grain growth and precipitation significantly influence the properties, while in the latter case the amount and nature of the martensite also contribute to the mechanical response. On the off-hand, however, ferrite grain growth and intragranular precipitation were shown to have a detrimental effect on toughness, ductility and bend ductility. It should also be noted that stabilizing elements, such as Ti and Nb, are far less effective during welding since the rapid cooling cycles tend to favor precipitation of Cr-rich precipitates. Thus, in fully ferritic microstructures, the nature of precipitation appears to have a strong influence on toughness and ductility, particularly when coupled with coarse grain size. The toughness and ductility of as-welded high-chromium high-carbon alloys is extremely poor due to the coupled effect of large grain size and high interstitial content. Unlike the low- and medium-chromium ferritic steels, weld properties are more sensitive to composition than microstructure and these alloys are extremely susceptible to HTE in the weld metal and HAZ. The use of carbon stabilizers (Ti and Nb) is effective in minimizing the deleterious effects of HTE, but excess stabilizer content has been also shown to have a

negative effect. In addition, stabilizers are only marginally effective in controlling grain growth and the embrittlement that results from this phenomenon. Due to the tendency for extreme grain growth, preheat and interpass temperature are to be avoided and minimized, respectively.

As it could be derived by what is stated so far, **high temperature embrittlement** is one of the most serious problems associated with the welding of the ferritic stainless steels. This form of embrittlement is a function of both composition and microstructure and is most damaging in the high-Cr alloys. High levels of interstitial elements, particularly carbon and nitrogen, have an additive effect. Large grain size also contributes to embrittlement, particularly in the HAZ. In the weld metal and HAZ, HTE can result in a dramatic loss in toughness and ductility relative to the base metal.

However, most information regarding the weldability of ferritic alloys is associated with either **hydrogen-induced cracking (HIC)** or **weld solidification cracking**. The low-chromium alloys tend to have low susceptibility to weld-related cracking, while the high-chromium alloys are generally welded under very controlled conditions to avoid cracking, or used in applications where welding is not required. The medium-chromium alloys have been reported to be susceptible to both HIC and weld solidification cracking. Weld solidification cracking, in particular, occurs during the final stages of freezing due to the combined effects of impurity and alloying element segregation, liquid film formation at grain boundaries, and imposed thermo-mechanical restraint. When the primary solidification phase is ferrite, weld solidification cracking susceptibility is generally low. All of the ferritic grades solidify as primary ferrite, and thus cracking in these steels is relatively rare. Addition of alloying elements (Ti and Nb) and high impurity levels will increase susceptibility to solidification cracking, since segregation during solidification can still lead to the formation of low-melting liquid films along solidification grain boundaries. Although not extensively reported, it is likely that HAZ liquation cracking is also possible in some alloys. Solidification cracking in Nb-bearing weld metals that were fully ferritic has also been reported, but the addition of nitrogen appeared to reduce the cracking susceptibility by promoting a finer solidification structure via the inoculation effect of Nb-rich nitrides. The addition of nitrogen also resulted in the formation of martensite in the weld metals.

Considering HIC, when martensite was present along ferrite grain boundaries, both the weld metal and HAZ were found to be crack susceptible. Cracking was often of the “delayed type”, requiring an incubation period for hydrogen diffusion, but proper preheat and PWHT conditions were found to eliminate this problem entirely.

5.3.3. Corrosion and Erosion Resistance^{[2] [25] [30] [44] [45] [46]}

Ferritic alloys have the lowest **cavitation erosion** among all stainless steels, but are generally susceptible to **crevice** and **pitting corrosion**. However, crevice and pitting corrosion can be fairly easily avoided by proper alloy selection and as such they are not considered to be a major problem for ferritic grades.

Moreover, due to the absence of nickel these grades exhibit significant resistance to **stress corrosion cracking (SCC)**, especially in chloride-containing environments, and are superior to most austenitic steels. That makes them perfect replacements for austenitic grades in certain occasions.

When generally encountered in various oxidizing environments, a protective oxide layer is likely to be formed on the metallic surface of ferritic grades. This layer is tight and adherent and it can prevent other aggressive elements in the environment from attacking and reacting with the steel. However, the layer can grow in thickness due to constant **oxidation**. The resulting porous layer will allow the gas to penetrate through to the base metal through pores or cracks. Silicon and aluminum are both beneficial for oxidation resistance. Low-thermal expansion and high-thermal conductivity of the ferritic base material can reduce changes in volume and thus spalling of the protective layer.

As a rule, ferrites perform better against **sulphide attacks** than austenites in oxidizing and reducing sulphurous atmospheres. In oxidizing atmospheres, attack can be delayed as long as the existing oxide scale is continuous and dense. However, scaling temperatures are up to 200°C lower than in air. Thus the oxide layer can grow faster and less compact, forming undesirable pores and cracks. With ferritic material in use, there is no risk of formation of low-melting-point nickel sulphides, which could destroy the remaining oxide layer and inhibit further passivation of the steel. In “reducing” sulphurous atmospheres, the oxygen activity may be sufficient to form a protective oxide layer, provided the Cr-content is higher than 25%. This usually being the case with these grades, ferritic steels are best utilized in these environments.

In terms of resistance to carburization, ferritic grades show less favorable results than austenitic ones due to their lack of nickel. Formation of chromium carbides or chromium nitrides, respectively, embrittles the material. Additionally, the surrounding matrix becomes chromium depleted and thus less able to form an oxide layer, which consequently reduces the scaling resistance of the material. Silicon has a beneficial effect on both **carbon** and **nitrogen pick-up**. Aluminum is only favorable in terms of carburization. The high nitrogen affinity of aluminum results in aluminum nitrides retarding formation of a protective alumina and leading to premature failure of the material.

However, when dealing with ferritic stainless steels, the major “player” in regard with corrosive attacks is **intergranular corrosion (IGC)**. Consensus opinion is that IGC results from a sensitization mechanism similar to that proposed for austenitic stainless steels. Cr-rich carbides and nitrides precipitate at the grain boundary, which results in a “Cr-depleted” zone that is susceptible to corrosive attack. As a result, resistance to IGC is a strong function of interstitial content. For high interstitial C + N levels intergranular precipitation cannot be suppressed even by rapid quenching and these alloys are inherently susceptible to IGC after welding. In alloys with low to moderate interstitial levels rapid cooling is effective in suppressing grain boundary precipitation and thus a critical cooling rate exists, over which sensitization will occur. Alloys with extremely low interstitial levels are the most resistant to IGC, but the critical interstitial level in high-chromium alloys must be extremely low to assure weld ductility. PWHT can significantly improve resistance of ferritic grades to IGC. Heating into the range of 700 to 950°C (1290 to 1740°F) is effective in “healing” the Cr-depleted regions. This treatment has been shown earlier to be effective in improving weld ductility and toughness. IGC resistance can also be improved through the addition of stabilizing elements, such as titanium and niobium. In summary, the IGC resistance of welds in ferritic stainless steels is controlled by a number of factors, including chromium and interstitial content and weld cooling rate.

5.4 SHIPBUILDING APPLICATIONS^{[25] [26] [27] [29] [30]}

Ferritic stainless steels are mainly in chemical processing. Consequently, that makes them ideal for use in chemical tankers.

The absence of nickel makes them ideal for use in chloride-containing environments, in replacement of austenitic alloys.

MARTENSITIC STAINLESS STEELS: Fe-Cr-C- (Ni-Mo) alloys

6.1 GENERAL ^{[2] [16] [25] [27]}

Martensitic stainless steels are based on the Fe-Cr-C ternary system. They were developed as a solution to the need of having a group of stainless alloys that exhibit average corrosion resistance and are hardenable by heat treating. As such they are straight chromium steels containing no nickel. Martensitic stainless steels are ferromagnetic and are mainly used in applications where hardness, strength and wear resistance are required.

These steels are also known as ***air hardening***, since cooling in still air after being withdrawn from a furnace as austenite is rapid enough to produce extensive martensite. Even weld cooling rates are sufficiently rapid to produce weld metal and HAZ microstructures that are predominantly martensitic. Martensitic stainless steels exhibit yield strengths ranging from 275 MPa in an annealed condition to 1900 MPa in a quenched and tempered condition. High hardness levels are also achievable, providing alloys with very good metal-to-metal wear resistance as well as excellent abrasion resistance. Their chemical composition (their almost complete lack of nickel) is the main reason martensitic alloys are not as good in corrosion resistance as other stainless steels, but they are ideal for applications where a combination of high strength and corrosion resistance under ambient atmospheric conditions is required. However, their low-chromium and low-alloying element content also makes them far less costly than the other types of stainless steel.

Martensitic stainless steels can be used in a fairly large range of temperatures, however they are not used above 650°C (1200°F), due to degradation of both

mechanical properties and corrosion resistance. However, during cooling after welding untempered martensite is formed. As such, martensitic alloys are considered to be the least weldable of all stainless steels.

Common applications of martensitic stainless steels include steam, gas, and jet engine turbine blades that operate at relatively low temperatures, steam piping, large hydro-electric turbines, fresh water canal locks, piping and valves for petroleum gathering and refining, and cladding for continuous caster rolls. High-chromium, high-carbon grades are also used to make items such as surgical instruments, cutlery, gears, and shafts.

Martensitic-austenitic stainless steels, or **supermartensitic** stainless steels, are obtained by increasing the nickel content slightly compared to the martensitic alloys. These alloys also often have slightly lower carbon content. However, the range of applications is largely that of martensitic stainless steels.

6.2 COMPOSITION AND MECHANICAL PROPERTIES ^{[2] [25] [27] [29] [30] [38]}

Based on their susceptibility to **hydrogen-induced cracking** or **cold cracking**, martensitic alloys are subdivided into three groups. The groupings are based on carbon content, since carbon content determines the hardness of the martensite in the as-welded condition, and martensite hardness is what influences cold cracking susceptibility. The first group consists of steels with about 0.06wt% C or less, which limits the maximum hardness to about 35 Rockwell C (HRC). Procedures used to weld low-alloy high-strength steels can also be used to weld steels of this group. Steels containing from 0.06 to about 0.30wt% C can be part of the second group, which exhibits an as-welded hardness that ranges from 35 HRC to about 55 HRC. The second group has greater cracking risk. The third group consists of steels with more than 0.30wt% C and an as-welded hardness of 55 to 65 HRC. This group calls for very specialized procedures for welding without cracking. In general, chromium content in martensitic stainless steels range from 11.5 to 18.0wt% and most of the structural grades contain carbon in the range of 0.10 to 0.25wt%. Some alloys also contain small additions of Mo, V, and W to provide high-temperature strength via the formation of stable carbides.

The following figure illustrates how alloying elements can determine the types of the martensitic group:

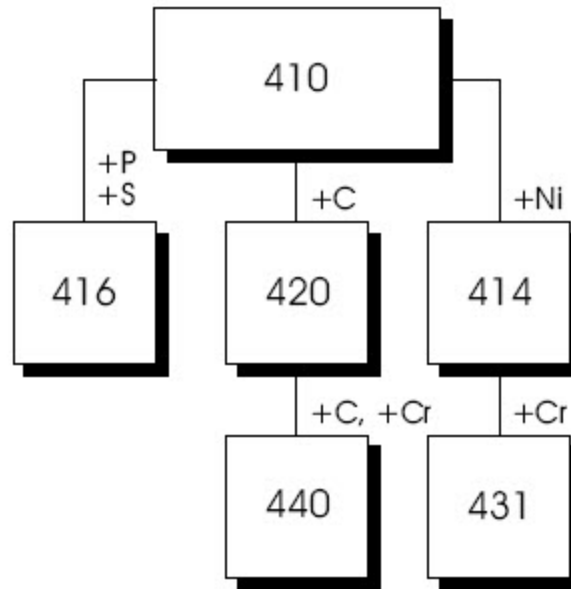


Figure 6.1 The Martensitic Group.

Type 410 is considered to be the basic martensitic grade and of the three basic stainless steels (types 304, 430, and 410) it contains the lowest alloy content. It is the least costly martensitic alloy and serves for general purpose. As all martensitic stainless steels it can be heat treated, but should only be used where corrosion is not severe (air, water, some chemicals, and food acids). Type 410S contains lower carbon than Type 410 and thus offers improved weldability, but lower hardenability. It is a general purpose corrosion and heat resisting chromium steel, that is often recommended for corrosion resisting applications. Type 414 has slightly raised nickel content for improved corrosion resistance, while Type 416 contains added phosphorus and sulphur for improved machinability. Increased carbon content is what improves the mechanical properties of Type 420, and Type 431 contains increased chromium for greater corrosion resistance and good mechanical properties. An even further increase in chromium and carbon content for Type 440 improves toughness and corrosion resistance.

The following table lists a general composition range of Standard Martensitic Stainless Steels:

Carbon:	0.15 to 1.20wt%
Manganese:	1 to 1.25wt%
Silicon:	0.5 to 1wt%
Chromium:	11.5 to 18wt%
Nickel:	-
Molybdenum:	0.6 to 2wt%
Phosphorus:	0.4wt%
Sulphur:	0.03 to 0.15wt%

More detailed compositions of Standard Martensitic Stainless Steels and Supermartensitic grades are shown on the following tables:

Type	UNS Number	Composition - Percent *							Other
		C	Mn	Si	Cr	Ni	P	S	
403	S40300	0.15	1.00	0.50	11.5-13.0		0.04	0.03	
410	S41000	0.15	1.00	1.00	11.5-13.0		0.04	0.03	
410Cb	S41040	0.18	1.00	1.00	11.5-13.5		0.04	0.03	0.05-0.3 Nb(Cb)
410S	S41008	0.08	1.00	1.00	11.5-13.5	0.8	0.04	0.03	
414	S41400	0.15	1.00	1.00	11.5-13.5	1.25-2.50	0.04	0.03	
414L		0.08	0.50	0.15	12.5-13.0	2.5-3.0	0.04	0.03	0.5 Mo; 0.03 Al
416	S41600	0.15	1.25	1.00	12.0-14.0		0.04	0.03	0.6 Mo
416Se**	S41623	0.15	1.25	1.00	12.0-14.0		0.06	0.06	0.15 min. Se
416 Plus X**	S41610	0.15	1.5-2.5	1.00	12.0-14.0		0.06	0.15 min.	0.6 Mo
420	S42000	0.15 min.	1.00	1.00	12.0-14.0		0.04	0.03	
420F**	S42020	0.15 min.	1.25	1.00	12.0-14.0		0.06	0.15 min.	0.6 Mo
422	S42200	0.20-0.25	1.00	0.75	11.0-13.0	0.5-1.0	0.025	0.025	0.75-1.25 Mo; 0.75-1.25 W; 0.15-0.3 V
431	S43100	0.20	1.00	1.00	15.0-17.0	1.25-2.50	0.04	0.03	
440A	S44002	0.80-0.75	1.00	1.00	16.0-18.0		0.04	0.03	0.75 Mo
440B	S44003	0.75-0.95	1.00	1.00	16.0-18.0		0.04	0.03	0.75 Mo
440C	S44004	0.95-1.20	1.00	1.00	16.0-18.0		0.04	0.03	0.75 Mo

Figure 6.2 Martensitic stainless steels composition table.

Type	Composition - Percent *								Other
	C	Mn	Si	Cr	Ni	Mo	Cu	N	
Lean	0.01	1.5	0.2	11	1.5	-	0.5	<0.01	-
Medium	0.01	0.5	0.2	13	4.5	1	0.5	0.05	Ti
High	0.01	0.5	0.2	12	6	2.5	0.2	0.05	Ti or V

Figure 6.3 Supermartensitic stainless steels composition table.

Since all properties of steels are largely determined by their microstructure, the composition of martensitic stainless steels is such that the austenite within them can transform into martensite. As a result, there is a degree of control on the mechanical properties by exploiting the phase change. It is considered common practice for martensitic stainless steels to be heat-treated through austenitisation at a temperature high enough to dissolve carbides, followed by quenching to obtain martensite. Given the high hardenability inherent in such alloys, the quench rate required to achieve martensite is not high; oil and water quenching are used only when dealing with thick sections.

Concerning the martensitic alloys' mechanical properties, one can clearly see in the following stress-strain curve that martensitic and martensitic-austenitic steels exhibit high yield and tensile strengths but low ductility, as opposed to austenitic steels that have excellent ductility but low yield strength.

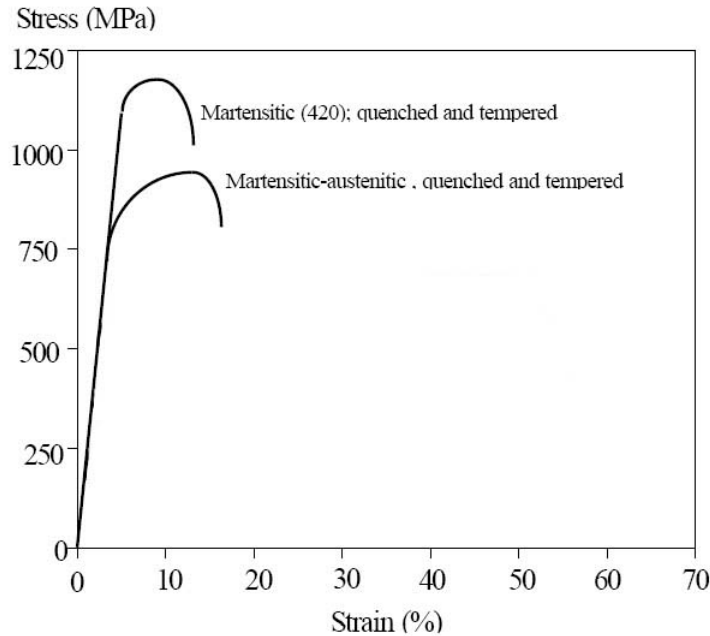


Figure 6.4 Stress-strain curve for a Type 420 martensitic and a martensitic-austenitic stainless steel.

Just as ferritic-martensitic steels, martensitic steels are characterized by their high strength and the fact that this strength is strongly affected by heat treatment. It's common practice to use martensitic steels in a hardened and tempered condition, and in this condition their strength increases with increasing carbon content. When chromium content exceeds 13wt% and carbon content exceeds 0.15wt% steels are completely martensitic after hardening. Lesser carbon increases ferrite content and therefore decreases the steel's overall strength. Martensitic steels have relatively low ductility, unlike ferritic-martensitic steels that exhibit high strength and good ductility in the hardened and tempered condition, despite their relatively low carbon content. They also possess excellent hardenability and even thick sections can be fully hardened, thus retaining their good mechanical properties even in thick sections.

Since the mechanical properties of martensitic stainless steels are heavily influenced by the heat treatments to which they are subjected, a point should be made to describe the general concept of these processes. The hardening treatment can be subdivided in two stages; the first stage involves the martensitic stainless steels being heated to the temperature range of 925 to 1070°C (620 to 1960°F) in order to produce an austenitic structure with carbon in solid solution. The steel's composition greatly influences the effect of austenitizing temperature and time on the steel's hardness and strength. Raising the austenitizing temperature will also improve hardness, but after a maximum temperature hardness will start decreasing again. Increased time at the austenitizing temperature, however, slowly reduces hardness. The second stage of the heat treatment involves quenching in oil or water depending on the steel grade. On cooling below the M_s - temperature, the starting temperature for the martensite transformation, the austenite transforms to martensite. The M_s - temperature lies within the range of 70 to 300°C (160 to 575°F), while the transformation is finished about 150 – 200°C below the M_s - temperature. Practically every alloying element will lower the M_s - temperature, however carbon has the greatest effect in this aspect. Essentially that means that higher alloyed martensitic grades are bound to contain

retained austenite in their microstructure, all because of the low temperature (below ambient) needed to finish the transformation of austenite to martensite.

In their hardened state martensitic stainless steels seem to favor strength and hardness over ductility and toughness. As such, in order to obtain useful engineering properties these alloys are usually tempered. The temperature used for tempering has an immense effect on the alloy's mechanical properties. Increasing the tempering temperature at just below 400°C (755°F) will cause a slight decrease in tensile strength and an increase in reduction of area. Hardness, elongation, and yield strength, however, are more or less unaffected by the tempering temperature. At around $450 - 500^{\circ}\text{C}$ ($845 - 935^{\circ}\text{F}$) there is a secondary hardening peak and as a result martensitic stainless steels tempered at that temperature exhibit an increase in yield strength, tensile strength and hardness. Above about 500°C there is a rapid reduction in strength and hardness, as well as a corresponding increase in both ductility and toughness.

Martensitic stainless steels are characterized by a transition in toughness, from tough to brittle, at a certain temperature. As shown on the following diagram, martensitic stainless steels have transition temperatures around or slightly below room temperature. However, martensitic alloys do exhibit a dip in the impact toughness curve at the temperature range of $450 - 500^{\circ}\text{C}$ ($845 - 935^{\circ}\text{F}$).

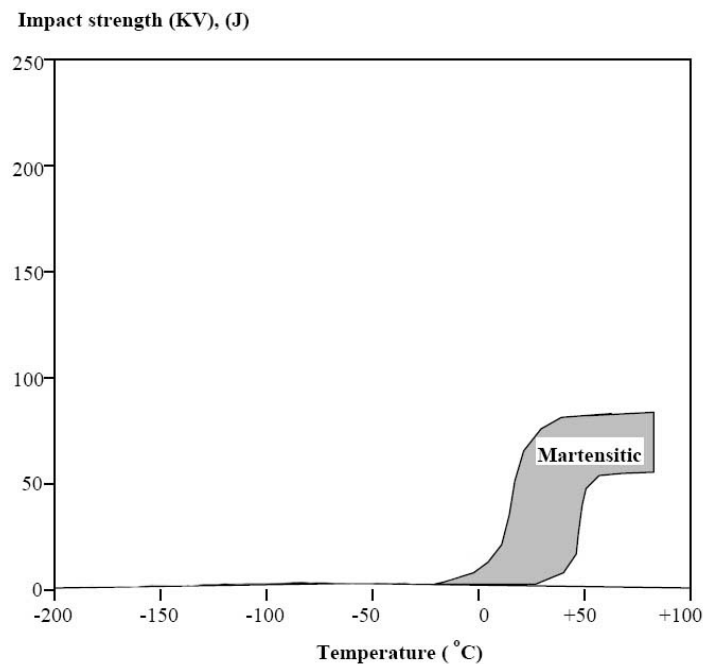


Figure 6.5 Impact toughness for martensitic stainless steels.

The high temperature strength of martensitic stainless steels is illustrated by the yield strength and creep rupture strength curves on the following diagram:

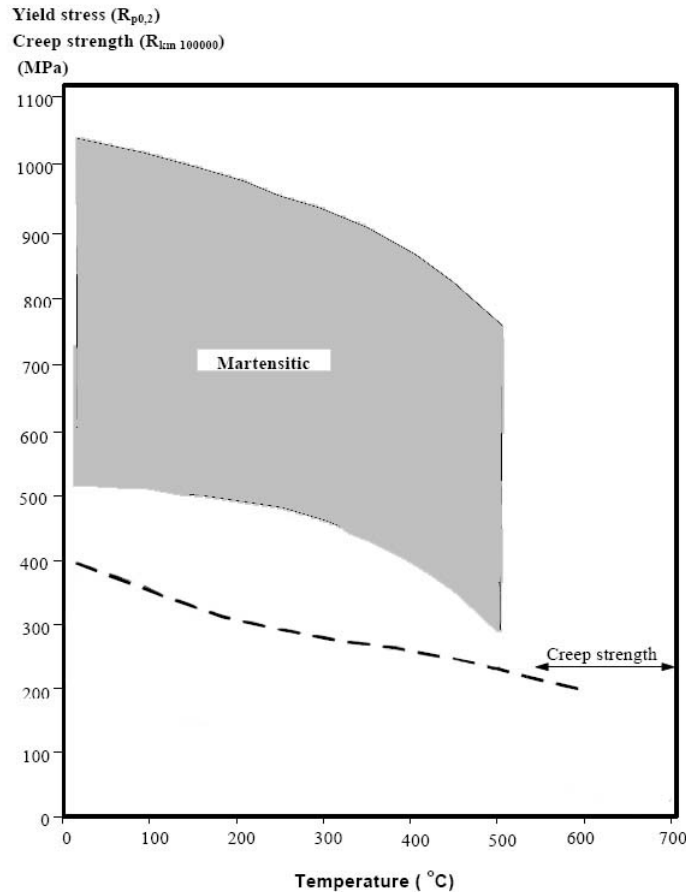


Figure 6.6 A high temperature strength diagram for martensitic stainless steels.

Martensitic and martensitic-austenitic steels in the hardened and tempered condition exhibit high elevated temperature strength at moderately elevated temperatures. However, the risk of over-tempering and embrittlement limits the useful upper service temperature and thus the creep strength is low. Martensitic alloys are usually used for applications that are subjected to temperatures up to 300°C (575°F), and only special grades are used at higher temperatures. The wide range of strength levels offered by different grades and heat treatments is responsible for the wide range of elevated temperature strength shown in the figure above.

The following table is derived from **ASM Metals Handbook, 8th edition, volume 1**, **ASM Metals Handbook, 9th edition, volume 3** and **ASTM standards** and synthesizes the mechanical properties of martensitic stainless steels:

Type	Condition	Tensile Strength		0.2% Yield Strength		Elong.	R.A.	Hardness
		Ksi	MPa	Ksi	MPa	%	%	Rockwell
403	Anneal	75	517	40	276	30	65	B82
403	*Temp. 800°F	195	1344	150	1034	17	55	C41
410	Anneal	75	517	40	276	30	65	B82
410	*Temp. 800°F	195	1344	150	1034	17	55	C41
410S	Anneal	60	414	30	207	22		B95M
410Cb	Anneal	70	483	40	276	13	45	
410Cb	*Temp. (Int.)	125	862	100	689	13	45	
414	Anneal	120	827	95	655	17	55	C22
414	*Temp. 800°F	200	1379	150	1034	16	58	C43
414L	Anneal	115	793	80	552	20	60	
416 Plus X	Anneal	75	517	40	276	30	60	
420	Anneal	95	655	50	345	25	55	B92
420	*Temp. 600°F	230	1586	195	1344	8	25	C50
422	Temp., Int.	140	965	110	758	13	30	
431	Anneal	125	862	95	655	20	60	C24
431	*Temp. 800°F	205	1413	155	1089	15	60	C43
440A	Anneal	105	724	60	414	20	45	B95
440A	*Temp. 600°F	260	1793	240	1655	5	20	C51
440B	Anneal	107	738	62	427	18	35	B96
440B	*Temp. 600°F	280	1931	270	1862	3	15	C55
440C	Anneal	110	758	65	448	13	25	B97
440C	*Temp. 600°F	285	1965	275	1896	2	10	C57

Figure 6.7 The mechanical properties of martensitic stainless steels.

6.3 PHYSICAL PROPERTIES^{[1] [2] [25] [29] [30] [37] [38]}

6.3.1. General^[37]

The main physical properties of martensitic stainless steels are as shown on the following table:

GRADES	Physical properties of stainless steels								
	DESIGNATIONS (▼)		DENSITY at 20°C [kg/dm ³]	MODULUS OF ELASTICITY at 20°C [GPa]	MEAN COEFFICIENT OF THERMAL EXPANSION [10 ⁻⁶ °K ⁻¹]		THERMAL CONDUCTIVITY at 20°C [W/(m·K)]	SPECIFIC HEAT at 20°C [J/(kg·K)]	ELECTRICAL RESISTIVITY at 20°C [Ω·mm ² /m]
	EN [N]	AISI/ASTM			20°C + 200°C	20°C + 400°C			
MARTENSITIC	1.4006 ^(b)	410	7,7	215	11,0	12,0	30	460	0,60
	1.4024 ^(b)		7,7	216	11,0	12,0	30	460	0,60
		414 ^(b)	7,78	200	10,4 ^(b)	11,0 ^(b)	25 ^(b)	460	0,70
	1.4005 ^(b)	416	7,7	215	11,0	12,0	30	460	0,60
	1.4029 ^(b)	416	7,7	215	9,9 ^(b)	11,0 ^(b)	30	460	0,55
	1.4021 ^(b)	420	7,7	215	11,0	12,0	30	460	0,60
	1.4028 ^(b)	420	7,7	215	11,0	12,0	30	460	0,65
	1.4031 ^(b)	420	7,7	215	11,0	12,0	30	460	0,55
	1.4034 ^(b)	420	7,7	215	11,0	12,0	30	460	0,55
		420F ^(b)	7,78	200	10,3 ^(b)	(n.r.)	25,2 ^(b)	460	0,55
	1.4116 ^(b)		7,7	215	11,0	11,5	30	460	0,65
		422 ^(b)	7,78	(n.r.)	11,2 ^(b)	11,4 ^(b)	23,9 ^(b)	460	(n.r.)
	1.4104 ^(b)		7,7	215	10,5	10,5	25	460	0,70
	1.4057 ^(b)	431	7,7	215	10,5	10,5	25	460	0,70
	1.4109 ^(b)	440A	7,7	215	11,0	11,5	30	460	0,65
	1.4112 ^(b)	440B	7,7	215	10,8	11,6	15	430	0,80
	1.4125 ^(b)	440C	7,7	215	10,8	11,6	15	430	0,80
	1.4110 ^(b)		7,7	215	11,0	12,0	30	460	0,62
	1.4419 ^(b)		7,7	215	11,0	12,0	30	460	0,62
	1.4122 ^(b)		7,7	215	10,8	11,6	15	430	0,80
	1.4313 ^(b)		7,7	200	10,9	11,6	25	430	0,60
	1.4418 ^(b)		7,7	200	10,8	11,6	15	430	0,80
	1.4422 ^(b)		7,7	200	10,8	11,6	16	450	0,75
	1.4423 ^(b)		7,7	200	10,8	11,6	16	450	0,75

Figure 6.8 The physical properties of martensitic stainless steels.

6.3.2. Weldability and Issues^{[2] [25] [30] [39]}

Methods commonly used with martensitic stainless steels are:

- Shielded Meta Arc Welding (SMAW)
- Gas Tungsten Arc Welding TIG (GTAW)
- Gas Metal Arc Welding MIG (GMAW)
- Flux-cored Arc Welding (FCW)

Martensitic stainless steels are considered to be the most difficult of the stainless steel alloys to weld. In martensitic and martensitic-austenitic stainless steels it is the quantity of martensite and its hardness that are mainly responsible for the weldability problems encountered. As stated before, the fully martensitic alloys are air-hardening, and are therefore extremely susceptible to hydrogen embrittlement. By welding at an elevated temperature (the steel's M_s temperature), the HAZ can be kept austenitic and tough throughout the entire welding process. Under normal weld cooling conditions, any austenite present at elevated temperature will transform to martensite. Many of the martensitic steels will also retain some high-temperature ferrite in the martensite matrix. The ferrite's presence is a function of the balance of ferrite-promoting to austenite-promoting elements. At higher carbon contents, however, the austenite phase field expands, enabling the formation of fully martensitic structures, thus

resulting in a much harder and more brittle martensite that is also more prone to hydrogen-induced cracking (a condition explained later) and possible brittle fracture. In contrast to these problems, when welding is finished and the weldment is allowed to cool, the mass of austenitic weld metal will transform to martensite. This transformation is followed by a volume expansion, because martensite is less dense than austenite, so the weld metal may arrive at ambient temperature under localized compressive stresses, thus reducing the risk of cracking.

Martensitic stainless steels in the as-welded condition have a HAZ than can exhibit a number of distinct microstructural regions. In a low-carbon alloy four distinct regions can be identified, as shown on the following picture:

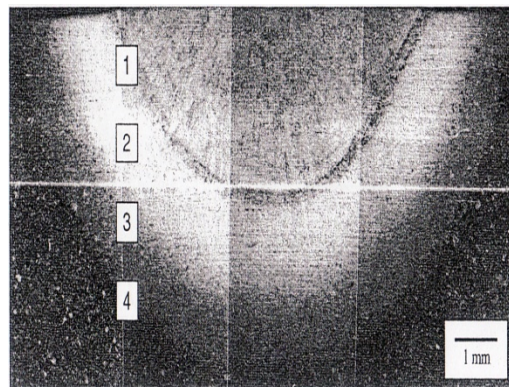


Figure 6.9 The four distinct regions encountered in the HAZ of a martensitic stainless steel in the as-welded condition.

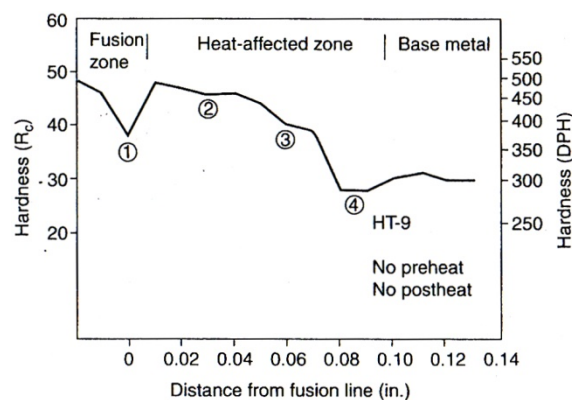


Figure 6.10 The four distinct regions as a function of hardness and distance from the fusion line.

Region 1 represents the portion of the HAZ just adjacent to the fusion boundary. The bulk of this region's microstructure at elevated temperature consists of austenite, even though some ferrite may also be present at the austenite grain boundaries. Being a strong austenite promoter, increasing carbon above 0.15wt% will increase the austenite phase field and decrease the amount of ferrite in the elevated-temperature microstructure. Upon cooling to room temperature, the austenite transforms to martensite and some of the ferrite remains in the microstructure, thus promoting local softening relative to the adjacent fusion zone and HAZ, as seen in the diagram right next to the picture that shows microhardness traverses of a low-carbon alloy. **Region 2** has a

fully austenitic microstructure at elevated temperature. Temperatures in this region are sufficiently high that the base metal carbides will dissolve and austenite grain growth will occur. Upon cooling, this region will be fully martensitic, since all or most of the carbon will have gone back into solution in the austenite. The peak HAZ hardness will generally occur in this region. **Region 3** is also heated into the austenite phase field during welding, but due to lower temperature than in region 2 austenite grain growth will not be so pronounced. Failure to dissolve the carbides results in a lower austenite carbon concentration and, as a consequence, a reduction in the hardness of the martensite formed upon cooling. **Region 4** shows little or no transformation to austenite and the microstructure appears virtually identical to the quenched and tempered base metal. Within the temperature range of 800 to 950°C (1470 to 1740°F) carbide coarsening can occur, resulting in some local softening relative to the base metal. However, if we are dealing with alloys having higher carbon content, the two-phase austenite + ferrite region will shrink and eventually disappear. This will result in the elimination of the softened region at the fusion boundary, since untempered martensite will extend all the way to the fusion boundary.

Another problem encountered, that is a function of ferrite content, is that if more than a few percent of ferrite exists in a predominantly martensitic stainless steel, the disparity in mechanical properties between the ferrite and the martensite needs to be taken into account. In hot-worked wrought steel, this is not so important, because the ferrite and martensite zones become aligned parallel to the rolling direction. As such, unless tensile strain is applied in the thickness direction, strain will be uniformly distributed in martensite and ferrite. In weld metal, however, ferrite areas are usually oriented perpendicular to the surface and, as a result, any strain applied parallel to the surface is concentrated disproportionately in the ferrite. This tends to cause fracture at both low stress and low elongation. This problem underlines the importance of predicting the presence of ferrite in martensitic stainless steel weld metals; the Balmforth is designed to predict the ferrite content in the weld metal of most martensitic stainless steels.

Postweld heat treatment (PWHT), after cooling, is almost always necessary for martensitic alloys. PWHT is used primarily to temper the martensite, but can also provide some stress relief when high level of residual stress are present. For a successful heat treatment the alloy should be tempered at about 480 to 750°C (895 to 1380°F), and preferably as a concluding heat treatment, although PWHT at temperatures as low as 200°C (390°F) are not unheard of. However, the weld must first be allowed to cool to below about 150°C (300°F). Practically, tempering at temperatures below 480°C (895°F) achieve no softening, but it can be used to improve toughness and dimensional stability after machining. Heat treatment durations depend on section thickness, but 30 minutes to 2 hours is normally sufficient.

Metallurgically, tempering promotes transformation of the martensite into ferrite and very fine carbides, thus reducing strength but improving ductility and toughness. However, if carbides other than chromium carbides form at intermediate temperatures, there may be a degree of secondary hardening that offsets the softening of the martensite. This allows lower-carbon weld metal, containing alloying elements besides chromium, to match the hardness and strength of higher-carbon base metal when used with a properly chosen PWHT. Care must be taken, however, not to temper higher-Cr alloys for excessive times, since sigma phase precipitation in the ferrite is possible, and that could cause the embrittlement of the structure. One should also take into consideration that in order to fully optimize properties relative to the base metal, the

entire structure should undergo a ***solution heat treatment (SHT)***, and then followed by quenching and tempering. This way the structure will be re-austenitized, and most or all of the ferrite will be dissolved in the weld metal and the HAZ. The direct result will be a uniform martensitic structure upon quenching. Unfortunately, this approach is usually not practical, due to size and/or logistical constraints.

As stated earlier, ***hydrogen-induced cracking*** is of major concern when dealing with martensitic stainless steels and a way to bypass this problem is by maintaining careful preheat and interpass control. Martensitic alloys containing less than 0.06wt% C, such as 13Cr/6Ni and 16Cr/5Ni/2Mo or Supermartensitic alloys, can often be welded without preheating or interpass control and even without postweld annealing, when in thin sections. In thin sections, if preheat and interpass control are considered a requirement they are set below the M_s temperature, to allow complete transformation to martensite and hydrogen diffusion. In thicker sections, however, preheat and interpass control should be above M_s to prevent possible cracking during fabrication. Steels of the 13Cr/4Ni type with low austenite content must be preheated to a working temperature of about 100°C (210°F), regardless of thickness. If optimal strength properties are desired, however, they can be heat treated at 600°C (1110°F) after welding. Following welding, the assembly is cooled slowly to room temperature to allow adequate time for hydrogen diffusion during the transformation process. It should be noted, however, that one should know the transformation temperatures of the base and filler metals that are being used in order to properly control preheat and interpass temperatures and avoid the unexpected presence of untempered martensite.

As a general guideline, the use of preheat and postweld heat treatments is recommended when welding these alloys, since these treatments also reduce any residual stresses. Moreover, low hydrogen processes reduce the amount of hydrogen uptake during welding and are considered essential when joining martensitic stainless steels. PWHT can also improve the toughness and ductility of martensitic alloys and this is considered critical, since materials with poor toughness and reduced ductility are prone to sudden catastrophic brittle failure. It is noteworthy that alloys containing a small vanadium addition soften more slowly as the PWHT temperature is raised, undoubtedly due to the secondary hardening effect of vanadium carbides present in the PWHT temperature range.

Another issue that should be taken into account when dealing with martensitic stainless steels is ***solidification cracking***. Even though martensitic steels solidify as ferrite, and hence are not very susceptible to solidification cracking, there are certain factors that are known to increase the likelihood of something like that actually happening. These factors include the presence of niobium in the alloy and very low Mn levels. On the other hand, very high carbon martensitic stainless steels may solidify as austenite, which renders them more sensitive to solidification cracking.

Reheat cracking is also detrimental for martensitic alloys, and usually occurs during the heating cycle imposed on a weld by a PWHT or residual heat from previous passes in multipass weldments. Molybdenum and impurities such as sulphur, phosphorus, antimony, tin, boron, and copper have been associated with reheat cracking in these steels. Minimizing any impurity levels, increasing heat input and eliminating stress concentrations can avoid the occurrence of reheat cracking. In general, however, martensitic stainless steels are not so susceptible to reheat cracking.

6.3.3. Corrosion and Erosion Resistance^{[25] [30] [40] [41] [42]}

Martensitic stainless steels, albeit extremely strong and tough, are not as corrosion resistant as the other stainless steel classes. In general, martensitic alloy corrosion resistance was never a real issue, since martensitic stainless steel has almost entirely been used for fabricating components requiring high mechanical properties and only moderate corrosion resistance.

Corrosion resistance for untreated martensitic stainless steels is provided solely by their thin passive film of chromium oxide. Although this layer can protect the material against low to moderate corrosive environments, it can be easily penetrated by more aggressive environments, leading to extensive **pitting** and **crevice corrosion**. In an acidic water solution containing 1% HCl the passive layer can be completely destroyed, leaving the underlying metal unprotected from active dissolution at a high rate.

For many of the applications martensitic alloys are being used for, nitriding proved to be very beneficial for their surface hardness and wear resistance, thus improving their performance and increasing their service life. Nitriding, however, also affects the corrosion properties of stainless steel, since nitrogen diffuses into the steel surface. As a matter of fact, nitriding austenitic stainless steel at high temperatures dramatically reduces their corrosion resistance due to the formation of chromium nitride and the depletion of free chromium in the steel matrix. As a result, the higher temperature nitrided stainless steels are no longer stainless. However, extensive studies on the effect of nitriding and nitriding temperature on the corrosion behavior of martensitic stainless steels conducted over the years have led to the conclusion that nitriding under certain conditions can actually be extremely beneficial to the corrosion properties of martensitic alloys. Even though chromium nitride precipitation occurred during nitriding at and above 460°C (860°F) and no precipitation was observed during nitriding at 420°C (790°F), electrochemical and immersion corrosion tests revealed that nitriding at any temperature improved corrosion resistance of the martensitic alloy, as shown on the following figures:

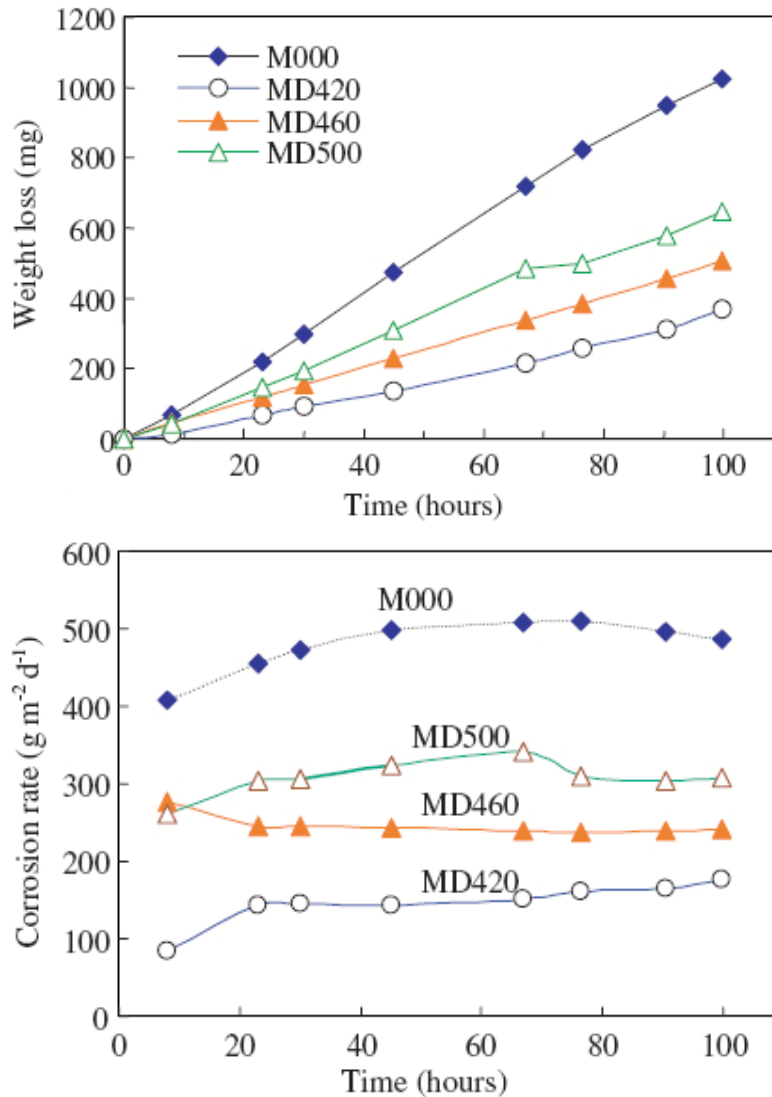


Figure 6.11 Nitriding at any temperature (420, 460, and 500°C) improves corrosion resistance of martensitic alloys.

The most probable explanation is that, since martensitic stainless steels contain higher amount of iron, higher amount of carbon and lower amount of chromium than austenitic stainless steels, a compound layer of iron nitride is more readily formed on martensitic alloys than on austenitic alloys during nitriding. Since the iron nitride formed on martensitic stainless steel has a similar structure as that formed on a low alloy steel surface, it could also possess good corrosion resistance and protect the underlying metal from corrosion. It should also be noted that nitrogen addition had the added side-effect of enhancing the repassivation ability of the steel's passive layer.

Despite their having the lowest corrosion resistance of all stainless steels, martensitic stainless steels do possess the highest cavitation erosion resistance, followed by austenitic and ferritic stainless steels. Heathcock et al. studied the cavitation erosion of various martensitic alloys and came to the conclusion that the cavitation erosion resistance of martensitic alloys actually increases with any increase in

their hardness. Carbon content and to a lesser degree, chromium content, seem to have a positive effect on martensitic alloys' cavitation corrosion resistance, since they both add to the hardness of the martensitic matrix. Carbon content, however, also has an adverse effect since it leads to the formation of coarse chromium carbides which provide sites for the initiation of erosion damage. Although even cast martensitic stainless steel has a fairly high cavitation erosion resistance, experiments revealed that **laser surface-melting (LSM)** could greatly increase cavitation erosion and pitting corrosion resistance. That is also evident on the following graphs (with the latter being a magnified version of the former for the more resistant specimens):

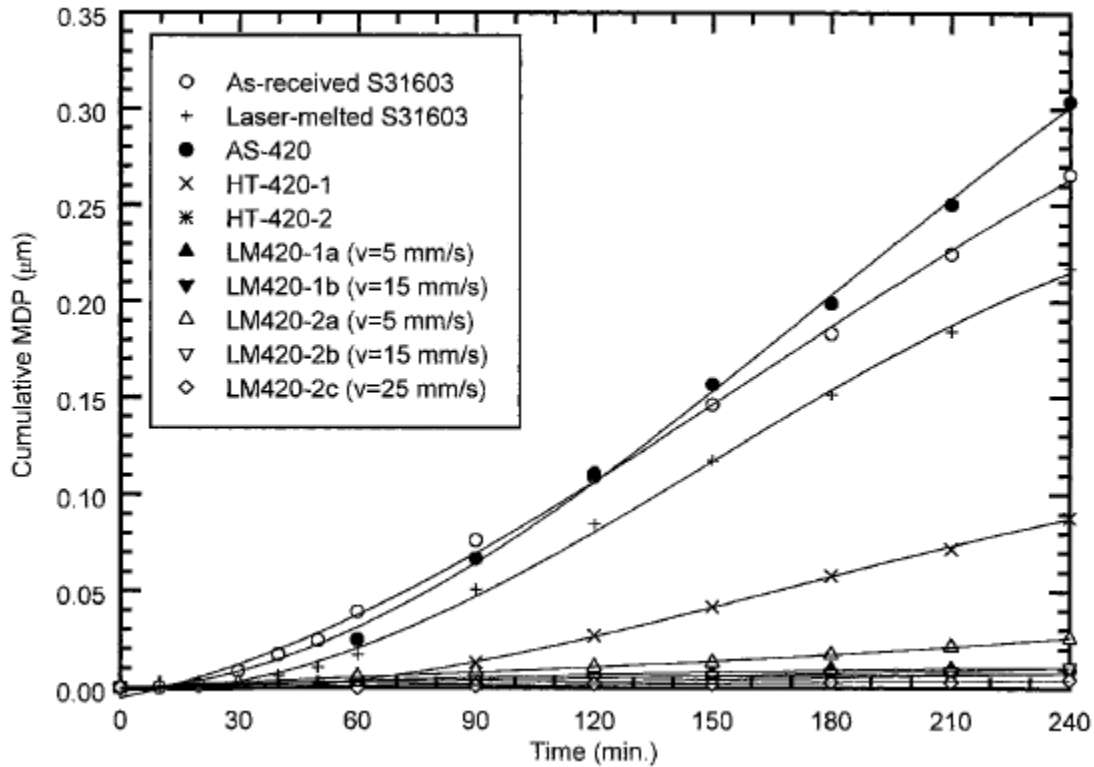


Figure 6.12 Laser surface-melting (LSM) improves cavitation erosion resistance of martensitic stainless steels.

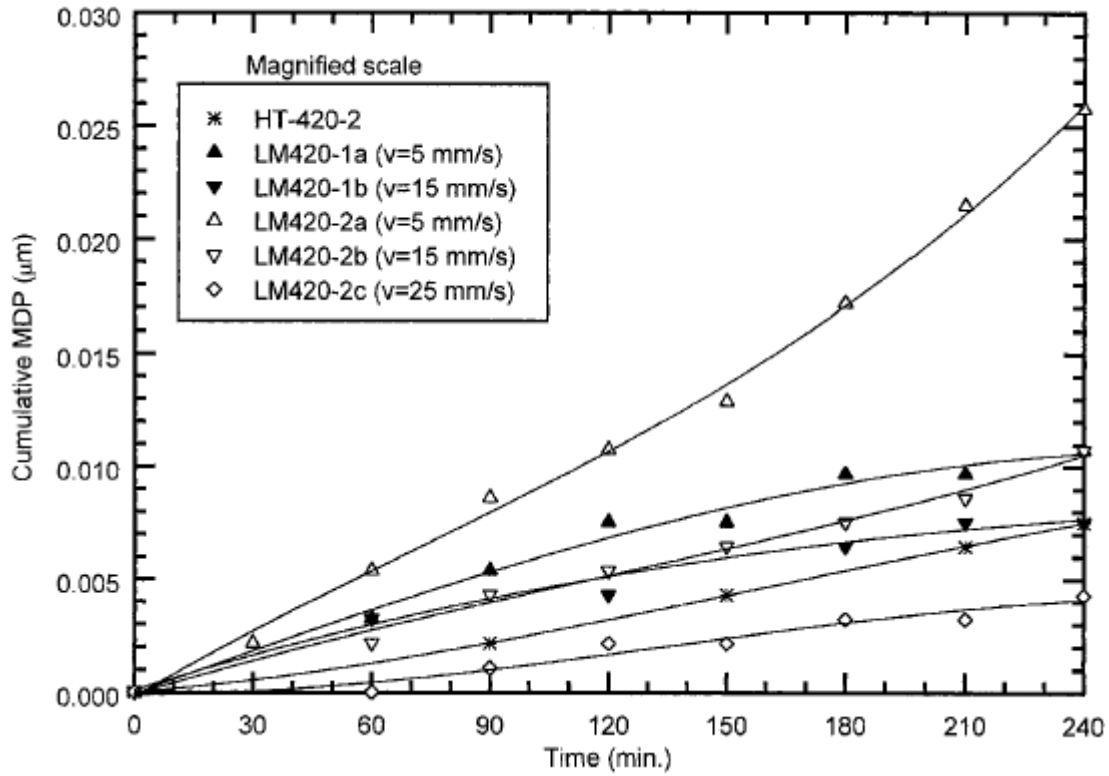


Figure 6.13 Laser surface-melting (LSM) improves pitting corrosion resistance of martensitic stainless steels.

6.4 SHIPBUILDING APPLICATIONS^{[25] [26] [27] [29] [30]}

Martensitic stainless steels are used in the processing of mildly sour gas and as such they can also be used in LNG carriers, wherever gas needs to be handled.

Supermartensitic steels are mainly found in pipes for offshore flowline applications.

DUPLEX STAINLESS STEELS: Fe-Cr-Ni-(Mo)- N alloys

7.1 GENERAL ^{[1] [2] [16] [25] [27]}

Duplex stainless steels derive their name from their room-temperature microstructure that is roughly half ferrite and half austenite, although in commercial alloys, the mix may be 60/40 respectively. Their development received more emphasis during periodic shortages of nickel, but it was not until recently that they were recognized as a separate family of stainless steels. Duplex steels are commonly used in applications requiring superior corrosion resistance, strength, or both.

Because of their higher ferritic content than that of austenitic steels, duplex steels are more ferromagnetic and exhibit higher thermal conductivity and lower thermal expansion. Duplex steels are also significantly stronger than their austenitic counterparts with typical yield strengths ranging above 425 MPa compared to 210 MPa for austenitic steels. However, higher strength means that duplex steels are also harder, possibly making them more attractive where abrasion is a concern. Because of their chemical composition ($C < 0.03\%$, $21.00\% < Cr < 26.00\%$, Mb up to 4.50%) duplex steels tend to form a number of embrittling precipitates at relatively low temperatures, and thus are not recommended for use at any temperature above 280°C (535°F). Duplex stainless steels also seem to undergo a ductile-to-brittle transition at low temperature making them unsuitable for service at cryogenic temperatures. In general, duplex stainless steels have a service temperature that is limited to a range of about -40 to 280°C (-40 to 535°F).

A perfect example of duplex stainless steel is type 2205. Type 2205 and type 2507, also known as “SuperDuplex” due to its higher corrosion resistance, are the most commonly used duplex stainless steels.

Duplex stainless steels are most commonly used in marine applications, petrochemical plants, desalination plants and generally seawater-bearing systems, heat exchangers and the papermaking industry.

7.2 COMPOSITION AND MECHANICAL PROPERTIES ^{[1] [2] [25] [29] [30] [33]}

As stated before, duplex stainless steels have been adjusted such that their base metal microstructure consists of 50% austenite and 50% ferrite. It should be pointed out however, that duplex stainless steels solidify as virtually 100% ferrite, and thus depend on partial solid-state transformation to austenite for their balanced microstructure. Such a transformation takes place during cooling through temperatures above approximately 1040°C (1900°F), when almost half of the ferrite transforms to austenite. This behavior can be accomplished by increasing Cr and decreasing Ni, as compared to austenitic stainless steels. Nitrogen is deliberately added to speed up the rate of austenite formation during cooling, and also to improve corrosion resistance. Molybdenum, tungsten, and/or copper are also added to some alloys to improve corrosion resistance. Because duplex stainless steels solidify as virtually 100% ferrite, they tend to have in general a higher ratio of ferrite- to austenite-promoting elements than do austenitic stainless steels.

Duplex steels are more prone than austenitic steels to precipitation of phases, and that causes embrittlement and reduced corrosion resistance. The formation of intermetallic phases, such as sigma phase, occurs in the temperature range of 600-950°C (1112-1742°F). The reformation of ferrite occurs in the range of 350-525°C (662-977°C). Exposures at these temperatures should therefore be avoided. In normal welding and heat-treatment operations the risk of embrittlement is low. However, certain risks do exist, for example in heat treatment of thick sections, especially if the cooling is slow.

More detailed compositions of Duplex Stainless Steels are shown on the following table:

Type	UNS Number	Composition - Percent *							ASTM A GRADE	Other
		C	Mn	Si	Cr	Ni	P	S		
Duplex Types										
2205	S32205	0.03	2.0	1.0	22.0	5.5	0.03	0.02		3.0 Mo; 0.18 N
2304	S32304	0.03	2.5	1.0	23.0	4.0				0.1 N
255		0.04	1.5	1.0	25.5	5.5				3.0 Mo; 0.17 N; 2.0 Cu
NU744LN		0.067	1.7	0.44	21.6	4.9				2.4 Mo; 0.10 N; 0.2 Cu
2507	S32750	0.03	1.2	0.8	25	5.5	0.035	0.020		4 Mo; 0.28 N

Figure 7.1 Duplex stainless steels composition table.

The main characteristic that differentiates duplex stainless steels from austenitic and ferritic stainless steels is that they have a higher yield strength and tensile strength. They are therefore often used in dynamically stressed machine parts. This characteristic is mainly derived from the fact that duplex steels contain higher levels of chromium and nitrogen. It should also be noted that the alloys containing even higher levels of chromium and nitrogen, the SuperDuplex alloys, exhibit even higher strength than the standard alloys, such as Type 2205. At low hot-working or annealing temperatures more austenite will be formed, at the expense of ferrite. However, a lower limit on these temperatures is imposed by the appearance of undesirable precipitates within the ferrite phase. Many of these precipitates are known to cause embrittlement to duplex alloys and are generally avoided. Moreover, the addition or increased levels of Cr, Mo and W tend to accelerate the formation of these precipitates, potentially causing implications with respect to embrittlement during post-weld heat treatment or multi-pass welding. These lower-temperature precipitation reactions effectively limit the use of duplex alloys to temperatures below 280°C (535°F).

The following stress-strain curve shows that duplex steels exhibit average yield and tensile strength and average ductility. Duplex steels have appreciably higher yield stress than both austenitic and ferritic steels, that increases with increasing carbon and nitrogen levels. Their strength will also increase, within limits, with increasing ferrite content. Their ductility is fairly good.

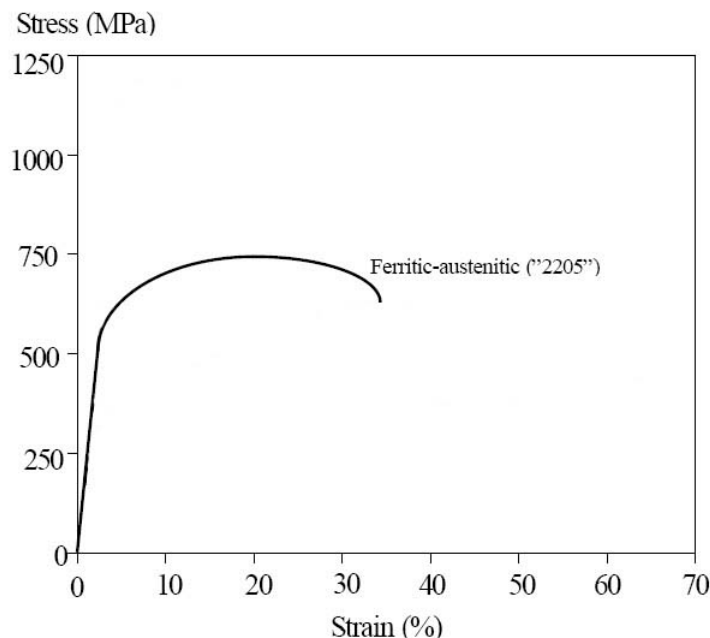


Figure 7.2 Stress-strain curve for the Type 2205 duplex stainless steel.

The general effect of cold work hardening is to increase the yield and tensile strengths and at the same time decrease the alloy's elongation. Although not as good as with austenitic steels, duplex steels exhibit strong work hardening.

Considering the fact that toughness increases with increasing temperature, the following impact strength diagram clearly shows that duplex steels tend to be relatively brittle at extremely low temperatures, but have a rather low transition temperature

where they exhibit a transition from brittle to tough behaviour. Increased ferrite content gives duplex steels a higher transition temperature, which effectively enlarges the temperature range for brittle behaviour. Duplex stainless steels have a transition temperature in the range of -60 to 0°C.

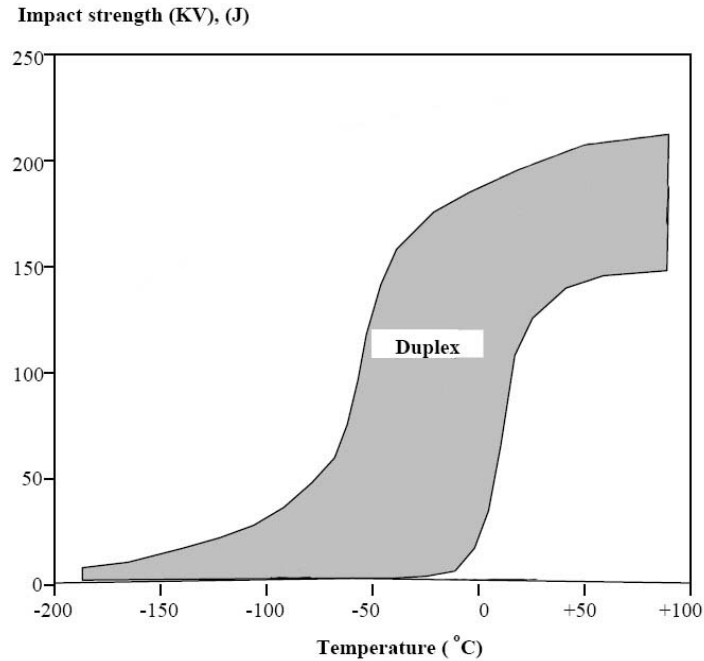


Figure 7.3 Impact toughness for duplex stainless steels.

As with Austenitic steels, duplex steels are also easily affected by cyclic loading at stress levels lower than their tensile strength. In general however, the high tensile strength of duplex stainless steels also implies high fatigue strength. The relation between duplex stainless steels tensile strength and fatigue strength can be seen at the following table:

Steel category	S_o/R_m Stress Ratio		Maximum stress
	R = -1	R = 0	
Duplex	0.55	0.35	yield strength

The following figure illustrates the high temperature strength of duplex stainless steels, where the dashed line shows the elevated temperature strength of some very high alloyed and nitrogen alloyed austenitic steels:

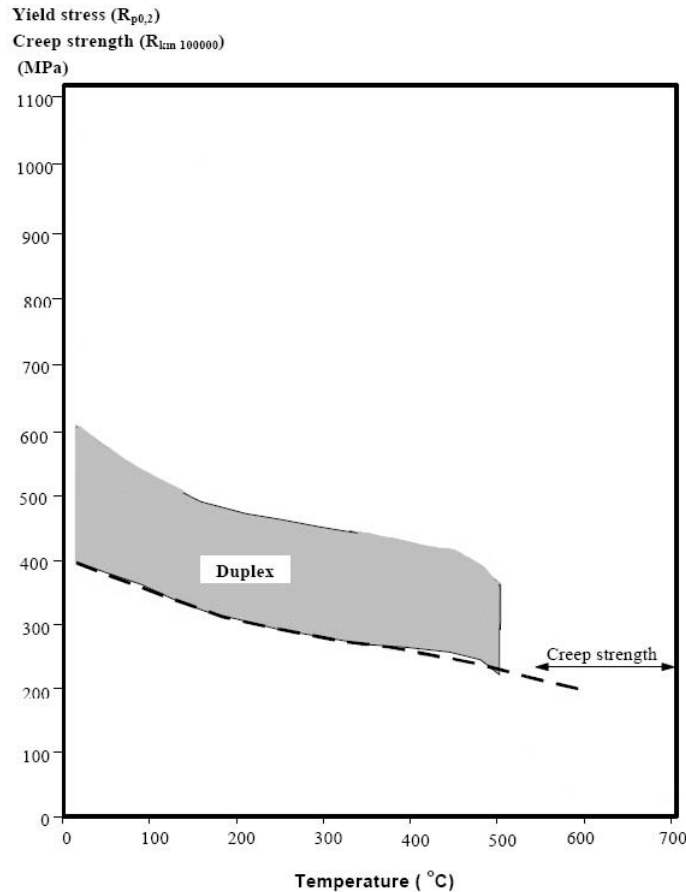


Figure 7.4 A high temperature strength diagram for duplex stainless steels.

By observation one can easily conclude that duplex steels exhibit greater yield stress than Austenitic steels and (as will be later proven) behave in the same way as the Ferritic steels, but have higher strength. However, their creep strength is relatively low and the upper service temperature limit is normally 350°C (662°F) due to the risk of embrittlement at higher temperatures.

The following table is derived from *ASM Metals Handbook, 8th edition, volume 1*, *ASM Metals Handbook, 9th edition, volume 3* and *ASTM standards* and synthesizes the mechanical properties of Duplex stainless steels:

Type	Condition	Tensile Strength		0.2% Yield Strength		Elong.	R.A.	Hardness
		Ksi	MPa	Ksi	MPa	%	%	Rockwell
Duplex Types								
2205		120	827	65	448	25		
2304		110	758	60	414	25		
255		110	758	80	552	15		
2507		116	800	80	550	15		

Figure 7.5 The mechanical properties of duplex stainless steels.

7.3 PHYSICAL PROPERTIES^{[1] [2] [25] [29] [30] [33] [34] [35] [36] [37]}

7.3.1. General^[37]

The main physical properties of duplex stainless steels are as shown on the following table:

GRADES	Physical properties of stainless steels								
	DESIGNATIONS (▼)		DENSITY at 20°C [kg/dm³]	MODULUS OF ELASTICITY at 20°C [GPa]	MEAN COEFFICIENT OF THERMAL EXPANSION		THERMAL CONDUCTIVITY at 20°C [W/(m·K)]	SPECIFIC HEAT at 20°C [J/(kg·K)]	ELECTRICAL RESISTIVITY at 20°C [Ω·mm²/m]
	EN [N°]	AISI/ASTM			[10⁻⁶·K⁻¹]				
			20°C + 200°C	20°C + 400°C					
DUPLEX	1.4460 ⁽⁹⁾	329	7,8	200	13,5	(n.r.)	15	500	0,80
	1.4462 ⁽⁹⁾	2205	7,8	200	13,5	14,0 ⁽⁹⁾	15	500	0,80
	1.4362 ⁽⁹⁾	2304	7,8	200	13,5	14,0 ⁽⁹⁾	15	500	0,80
	1.4655 ⁽⁹⁾		7,8	200	13,5	(n.r.)	15	500	0,80
	1.4410 ⁽⁹⁾	2507	7,8	200	12,5 ⁽⁹⁾	13,5 ⁽⁹⁾	15	500	0,80
	1.4507 ⁽⁹⁾	255	7,8	200	12,5 ⁽⁹⁾	13,5 ⁽⁹⁾	15	500	0,80
	1.4501 ⁽⁹⁾		7,8	200	13,5	(n.r.)	15	500	0,80
	1.4424 ⁽⁹⁾		7,8	200	13,5	(n.r.)	13	475	0,80
	1.4477 ⁽⁹⁾		7,7	200	12,0	(n.r.)	13	470	0,80

FERROMAGNETIC									
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Figure 7.6 The physical properties of duplex stainless steels.

7.3.2. Weldability and Issues^{[2] [25] [30] [33] [35]}

Methods commonly used with duplex stainless steels are:

- Shielded Meta Arc Welding (SMAW)
- Gas Tungsten Arc Welding TIG (GTAW)
- Gas Metal Arc Welding MIG (GMAW)
- Flux-cored Arc Welding (FCW)
- Plasma Arc Welding (PAW)
- Submerged Arc Welding (SAW)
- Laser
- Resistance Welding
- High Frequency Welding

Today's duplex stainless steels have considerably better weldability than earlier grades. However, since welds and heat-affected zones are rapidly cooled from temperatures near the ferrite solvus, duplex stainless steels tend to present much higher ferrite content in the weld metal and HAZ than in the base metal. As a consequence, it has become common practice to intentionally add nitrogen in duplex stainless steels, usually a content ranging between 0.08 and 0.35wt%, as a key element for lowering the effective quench temperature so that the HAZ phase balance can approach that of the duplex stainless steel base metal, effectively making nitrogen content the key to manipulating the phase balance under weld cooling conditions.

Nitrogen content is also used as a means of improving strength and pitting corrosion resistance. When the microstructure is balanced, nitrogen partitions between the ferrite and austenite and remains predominantly in solid solution. However,

nitrogen is much more soluble in austenite than it is in ferrite. As a result, in cases where the ferrite content is high, such as in weld metal and HAZ under rapid cooling conditions, an intense nitride precipitation reaction occurs upon cooling since the solubility limit of the ferrite is exceeded and nitrogen has insufficient time to partition to the austenite. The consequences of this intense precipitation are loss of ductility, toughness and corrosion resistance.

On another note, time above the ferrite solvus temperature has a dramatic enlarging effect on ferrite grain growth, since above this temperature there is effectively no impediment to grain growth. Since ferrite grain size has a detrimental effect on toughness and ductility, it is generally advisable to minimize the time in the fully ferritic region and this can be achieved by either controlling the composition or the weld heat input and thermal conditions. A decrease in the Cr_{eq}/Ni_{eq} ratio can increase the ferrite solvus temperature, thus decreasing the time above this temperature. On the other hand, decreasing the heat input promotes steeper thermal gradients and minimizes the time in the fully ferritic region.

One other thing that is connected with rapid cooling conditions, such as in duplex stainless steels weld metals and HAZs, is the fact that the ferrite-austenite phase balance tends to be higher than equilibrium. As a result, reheating of the weldment, such as during multipass welding, can result in further growth of existing austenite or nucleation of new austenite; a condition that is termed **secondary austenite precipitation** and can significantly alter the ferrite-austenite phase balance. Secondary austenite precipitation can improve the toughness of deposits that would otherwise have high ferrite contents and there has been some evidence that it may reduce pitting corrosion resistance.

As shown on the following figure, practically all duplex alloys solidify in the F mode and have inherently higher solidification cracking susceptibility than the austenitic alloys that solidify in the FA mode. This is caused by the presence of ferrite-ferrite boundaries that are much easier to wet by liquid films than ferrite-austenite boundaries. In practice, however, duplex steels are actually quite resistant to **weld solidification cracking**, probably due to the fact that their low impurity levels minimize the likelihood of continuous grain boundary liquid films.

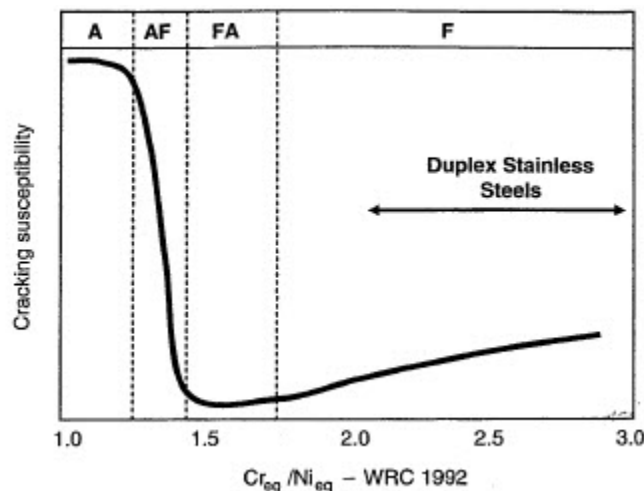


Figure 7.7 Cracking susceptibility of duplex stainless steels, as a function of their composition.

Duplex steels are considered to be resistant to **hydrogen-induced cracking**. However, there have been numerous reports on failures occurring in duplex stainless steel weldments due to a combination of high hydrogen contents and poor microstructure control. In order to prevent this SMAW and SAW processes should be practiced with low-hydrogen, while hydrogen additions to argon gas shielding for GMAW and GTAW processes should be avoided.

Due to the high Cr and Mo contents of duplex alloys, precipitation of these phases is likely if they are subjected to the appropriate temperature range, thus making duplex steels susceptible to **intermediate-temperature embrittlement**. Formation of these intermetallic phases is detrimental to ductility, toughness, and corrosion resistance.

In general, annealing of weldments usually reduces the yield strength of the joint. However, base metal and welding filler metal do not include requirements for toughness. As such, the use of duplex steels in certain applications can be a perfect method of controlling toughness. The possibility of high ferrite content in the weld metal and HAZ of these alloys allow them to behave in the same manner as ferritic stainless steels, exhibiting a ductile-to-brittle transition temperature in both base metal and weldments. In the offset, where a toughness requirement is actually imposed for the HAZ, high nitrogen content in the base metal is once more the answer to meeting this requirement. Weld metal toughness is also affected by oxygen content and welding processes that produce weld metal of lower oxygen content, such as low-oxygen GMAW and GTAW, generally produce weld metal of higher toughness.

7.3.3. Corrosion and Erosion Resistance^{[2] [25] [30] [33]}

The duplex steels cover a wide range of corrosion performance in various environments. As far as **uniform corrosion** is concerned, due to their high chromium content, duplex alloys offer excellent resistance in many media. For example, Type 2205 shows very good resistance in sulphuric acid contaminated by chloride ions, similar to that of 904L Austenitic stainless steel. Most duplex alloys do not seem to handle extremely well in hydrochloric acid, but certain high-alloyed SuperDuplex alloys, such as 2507 and to some extent even 2205, can be used in dilute hydrochloric acid. Phosphoric acid almost always contains corrosive contaminations, e.g. in the form of chlorides and fluorides, but still duplex steels offer very good resistance even in acids that have fairly high halide content. In strongly oxidizing acids, such as nitric acid, non-molybdenum alloyed steels are more resistant than the molybdenum alloyed ones. As such, certain duplex alloys are good for applications in such environments, due to their high chromium content in combination with low molybdenum content.

The amount of chromium, molybdenum and nitrogen in the steel is responsible for the steel's resistance to **pitting** and **crevice corrosion**. Due to their different alloying levels, various duplex steels show considerable differences in this respect. Several pitting resistance equivalent (PRE) formulae have been developed to represent the fact that pitting resistance for stainless steels is composition dependent. The following is the most popular:

$$PRE_N = Cr + 3.3 \cdot (Mo + 0.5W) + 16N$$

PRE_N values greater than 40 are loosely taken to identify an alloy as SuperDuplex. However, since microstructure can also influence pitting corrosion resistance, PRE

values cannot be used as effectively with weld metals and HAZs, since they tend to have unbalanced microstructures and significant precipitation.

Duplex stainless steels are also much less sensitive to **stress corrosion cracking** in chloride environments than austenitic steels, mostly due to their low-Ni and high-Cr contents as well as a continuous ferritic phase. The SuperDuplex alloy 2507 in particular shows no susceptibility to SCC under 300°C (570°F) and certain Cl⁻ ion concentrations.

The 2205 and 2507 alloys have also demonstrated good resistance to **sulphide stress corrosion cracking**; a type of stress corrosion cracking encountered in the presence of hydrogen sulphide and chlorides. As a matter of fact, both 2205 and 2507 alloys are approved materials according to NACE MR0175 “Standard Material Requirements – Metals for Sulfide Stress Cracking and Stress Corrosion Cracking Resistance in Sour Oilfield Environments”, with official statements to verify that.

The combination of high mechanical strength and very good resistance to corrosion has given duplex steels high corrosion fatigue strength. As shown on the following S-N curves, the corrosion fatigue strength of 2205 alloy is considerably higher than that of 4436.

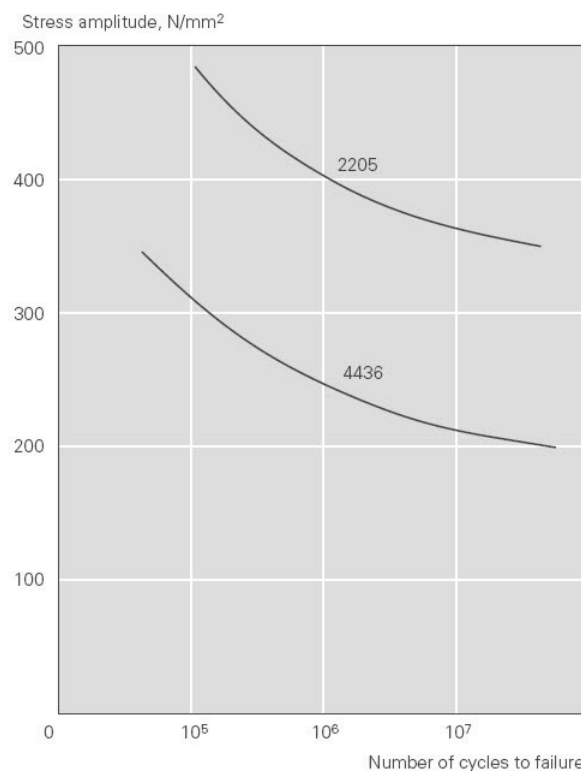


Figure 7.8 Corrosion fatigue strength of Type 2205 duplex stainless steel, as opposed to Type EN 1.4436 Outokumpu austenitic stainless steel.

Intercrystalline corrosion also seems to pose no threat to duplex steels, because of their duplex microstructure and low carbon content. Due to its composition, austenite is reformed in the HAZ after welding and the risk of undesirable precipitation of carbides and nitrides in the grain boundaries is minimized.

Last but not least, duplex steels are especially good against ***erosion corrosion***, thanks to their combination of high surface hardness and good corrosion resistance. This makes duplex steels especially good for applications where systems are subjected to particles causing hard wear, e.g. pipe systems containing water with sand or salt crystals.

7.4 SHIPBUILDING APPLICATIONS^{[25] [26] [27] [29] [30]}

Some of the uses of Duplex stainless steels are the following:

- Chemical tankers
- Waterjet propulsion systems
- Pressure vessels
- Large storage tanks
- Firewalls and blast walls on offshore platforms
- Cargo tanks and pipe systems in chemical tankers
- Fuel-gas cleaning
- Seawater systems

PRECIPITATION-HARDENING STAINLESS STEELS

8.1 GENERAL ^{[1] [2] [16] [27]}

The first commercial precipitation-hardening stainless steel is attributed to Smith et al. and its composition was about 17wt% Cr, 7wt% Ni and 0.7wt% Ti. In this alloy titanium was the alloying element responsible for the precipitation hardening reaction. Not long after that aluminum, copper, and beryllium were also found to produce similar effects. However, beryllium was soon abandoned due to health concerns regarding its use in commercial alloys.

Precipitation-hardening stainless steels are designated as such because they derive a significant part of their strength from precipitation reactions. They are grouped together because they contain elements that will form fine precipitates when heat treated. The formation of martensite in these steels is also another major factor in their hardening. These steels are subclassified by the predominant microstructure that constitutes the alloys as ***martensitic***, ***austenitic*** and ***semiaustenitic***.

Precipitation-hardening alloys are capable of achieving high tensile strengths, over 1520 MPa (220 ksi) in some of their grades. In addition to their strength they also exhibit good ductility and toughness, should they be properly heat treated. Service temperatures are typically limited to about 315°C (600°F) for continuous service. Austenitic PH grades may be used at up to 650°C (1200°F) or even higher. Most of PH stainless steels have corrosion resistance approaching that of Type 304 austenitic stainless steel. However, since any heat treatment can potentially have great effect on corrosion resistance, heat treatment conditions should be carefully controlled when corrosion is an issue. In general, when high specific strength and corrosion resistance are required, these alloys are almost certain to stand up to the occasion and meet the needs of the application at hand.

Despite the high strength and corrosion resistance, however, the PH stainless steels are not as widely used as the other grades, mainly because many of these alloys are martensitic and precipitation strengthened and they are more difficult to fabricate than other stainless steels, usually also requiring special heat treatments. Because of their processing requirements and low-volume use, these alloys are generally more expensive to apply than other stainless steel grades, but common applications for the martensitic grades include valves, gears, splines, and shafts. Semiaustenitic PH alloys are used in pressure vessels, aircraft frames, and surgical instruments. Austenitic PH grades are used in jet engine frames, hardware, and turbine blades.

8.2 COMPOSITION AND MECHANICAL PROPERTIES ^{[2] [25] [27] [29] [30]}

The PH stainless steels are not classified with the standard numbering system used by other stainless steels. Instead they are known by generally used trade names and they can also be identified by their UNS number or ASTM type.

Most of the PH alloys are martensitic, or a mixture of martensite, ferrite, and small amounts of austenite. PH grades, in general, offer the designer a unique combination of fabricability, strength, ease of heat treatment, and corrosion resistance not found in any other class of material. These grades include the widely known 17-4PH (17Cr – 4Ni) and 15-5PH (15Cr – 5Ni) alloys.

The austenitic precipitation-hardenable alloys have, to a large extent, been replaced by the more sophisticated and higher strength **superalloys**. The martensitic precipitation-hardenable stainless steels are really the workhorse of the entire PH family. While designed primarily as a material to be used for bar, rods, wire, and forgings, martensitic precipitation-hardenable alloys are beginning to find more use in the flat rolled form. The same goes for the semiaustenitic precipitation-hardenable stainless steels, primarily designed as a sheet and strip product, which have found many applications in other product forms.

Developed primarily as aerospace materials, many of these steels are gaining commercial acceptance as truly cost effective materials, since it has been estimated that almost 2% of a steel construction's budget is spent dealing with corrosion.

The following table lists detailed compositions of PH stainless steels:

Type	UNS	Composition - Percent *							ASTM GRADE	Other
	Number	C	Mn	Si	Cr	Ni	P	S		
Precipitation-Hardening Types										
PH 13-8 Mo	S13800	0.05	0.10	0.10	12.25-13.25	7.5-8.5	0.01	0.008		2.0-2.5 Mo; 0.90-1.35 Al; 0.01 N
15-5 PH	S15500	0.07	1.00	1.00	14.0-15.5	3.5-5.5	0.04	0.03		2.5-4.5 Cu; 0.15-0.45 Nb(Cb) + Ta
17-4 PH	S17400	0.07	1.00	1.00	15.5-17.5	3.0-5.0	0.04	0.03	630	3.0-5.0 Cu; 0.15-0.45 Nb(Cb) + Ta
17-7 PH	S17700	0.09	1.00	1.00	16.0-18.0	6.5-7.75	0.04	0.03	631	0.75-1.15 Al
PH 15-7 Mo	S15700	0.09	1.00	1.00	14.0-16.0	6.5-7.75	0.04	0.03		2.0-3.0 Mo; 0.75-1.5 Al
17-10 P		0.07	0.75	0.50	17.0	10.5	0.28			
A286	S66286	0.08	2.00	1.00	13.5-16.0	24.0-27.0	0.040	0.030	660	1.0-1.5 Mo; 2 Ti; 0.3 V
AM350	S35000	0.07-0.11	0.5-1.25	0.50	16.0-17.0	4.0-5.0	0.04	0.03		2.5-3.25 Mo; 0.07-0.13 N
AM355	S35500	0.10-0.15	0.5-1.25	0.50	15.0-16.0	4.0-5.0	0.04	0.03		2.5-3.25 Mo
AM363		0.04	0.15	0.05	11.0	4.0				0.25 Ti
Custom 450	S45000	0.05	1.00	1.00	14.0-16.0	5.0-7.0	0.03	0.03		1.25-1.75 Cu; 0.5-1.0 Mo 8 x %C - Nb(Cb)
Custom 455	S45500	0.05	0.50	0.50	11.0-12.5	7.5-9.5	0.04	0.03		0.5 Mo; 1.5-2.5 Cu; 0.8-1.4 Ti; 0.1-0.5 Nb(Cb)
Stainless W	S17600	0.08	1.00	1.00	16.0-17.5	6.0-7.5	0.04	0.03		0.4 Al; 0.4-1.2 Ti

Figure 8.1 PH stainless steels composition table.

It should be noted at this point that there is serious doubt as to whether or not the alloys S35000 and S35500 should be considered as PH alloys, since neither contains copper, niobium, aluminum, or titanium. They obtain some hardening from nitride precipitation, but their strength does not increase due to precipitation.

In the case of ***Martensitic Precipitation-Hardening stainless steels***, Brooks and Garrison showed that some PH stainless steels (namely 13-8Mo, 15-5PH and Custom 450) solidify as essentially 100% ferrite. During cooling the ferrite transforms almost entirely to austenite. Whatever ferrite persists will remain to ambient temperatures, with no transformation at all. Wrought steel will have the ferrite aligned in directions parallel to the rolling direction (parallel to the surface) and this distribution of ferrite, if present, can result in markedly lower strength and ductility in the through-thickness direction compared to properties in directions parallel to the rolled surface.

A homogenization treatment can always be used as part of the annealing process when the steel is in the austenitic phase and this homogenization tends to reduce ferrite content. The austenite subsequently transforms to martensite on cooling to temperatures approaching ambient, however, a small amount of austenite may be retained to ambient temperature or even below. Once a martensitic PH stainless steel has been transformed to largely martensite, an aging treatment can be performed to precipitate a hardening phase and increase the strength above that of the freshly formed martensite. The aging temperature for maximum strength depends on the specific precipitate being produced. The aged condition is generally indicated by the letter "H" followed by the temperature in degrees Fahrenheit at which the aging took place. The precipitates that harden these steels depend on the alloying elements present.

Sometimes, during the aging treatment, at least with some alloys, austenite reversion can take place. This reverted austenite, distinct from the retained austenite, was found to be enriched in Ni and Mn, and to be very resistant to martensite transformation. This austenite has much lower strength than that of martensite and is practically responsible for some alloys having lower strength with increasing aging temperature, something that was at first falsely attributed to precipitate coarsening and loss of precipitate coherency with the matrix.

When dealing with ***Semi-Austenitic Precipitation-Hardening stainless steels***, as with martensitic PH stainless steels, their solidification is as primary ferrite. Transformation to austenite occurs at high temperatures, but some ferrite is retained to ambient temperatures in some alloys (namely 17-7PH, 15-7Mo, AM350, and AM355). Solution annealing at 1040 to 1065°C (1900 to 1950°F) causes some of the ferrite to transform to austenite and homogenizes the austenite. Cooling to ambient temperature from the solution anneal leaves the austenite largely untransformed and this is commonly referred to as ***condition A***, a condition in which the steel is relatively soft and ductile. In this condition the steel can be easily machined, cold-formed, or welded. Cold forming, however, may induce some martensite transformation.

For the semi-austenitic PH stainless steels, the hardening treatment is more complex than for the martensitic PH stainless steels, achieved by a heat treatment traditionally consisting of three steps. It should be noted, however, that as with martensitic PH alloys, the semi-austenitic PH stainless steels also appear to experience austenite reversion during heat treatment at the higher aging temperatures. They also experience a lowering in strength after these higher aging temperatures, as with martensitic PH grades, so the conclusion of Hochanadel et al.,

that austenite reversion and not precipitate coarsening or loss of coherency explains the lower strength, is probably valid for the semi-austenitic stainless steels as well.

Lastly, in the case of **Austenitic Precipitation-Hardening stainless steels**, their physical metallurgy is much simpler than that of martensitic and semi-austenitic PH stainless steels. There is no transformation to martensite that needs to take place, and there is no austenite conditioning treatment. The austenite PH stainless steels solidify as austenite and remain so at all temperatures.

In the condition provided in from mills, austenitic PH alloys are quite soft and the typical strength for commercial product is expected to be about 275 MPa (40 ksi). The austenitic PH stainless steel can be aged in the approximate temperature range of 675 to 760°C (1250 to 1400°F) to become hardened. The aging time is much longer than for martensitic and semi-austenitic PH stainless steels, reaching from 16 to 20 hours, apparently because diffusion of alloy elements to form precipitates is much slower in the austenite matrix than it is in the martensite matrix of the latter alloys. Something that should be noted, however, is that the fully hardened strength level of the austenitic PH stainless steels is considerably less than that of the martensitic or semi-austenitic PH stainless steels.

The following table is derived from **ASM Metals Handbook, 8th edition, volume 1**, **ASM Metals Handbook, 9th edition, volume 3** and **ASTM standards** and synthesizes the mechanical properties of PH stainless steels:

Type	Condition	Tensile Strength		0.2% Yield Strength		Elong.	R.A.	Hardness
		Ksi	MPa	Ksi	MPa	%	%	Rockwell
Precipitation Hardening Types								
Ph13-8 Mo	H950	220	1517	205	1413	8	45	C45
15-5PH	H900	190	1310	170	1172	10	35	C44
15-5PH	H1150	135	931	105	724	16	50	C32
17-4PH	Sol. Ann.	150	1034	110	758	10	45	C33
17-4PH	H900	200	1379	178	1227	12	48	C44
17-7PH	Sol. Ann.	130	896	40	276	35		B85
17-7PH	RH950	235	1620	220	1517	6		C48
PH15-7 Mo	Sol. Ann.	130	896	55	379	35		B88
PH15-7 Mo	RH950	240	1655	225	1551	6	25	C48
17-10P	Sol. Ann.	89	613	37	255	70	76	B82
17-10P	H1300	143	996	98	676	20	32	C32
A286	H1350	130	896	85	586	15		
AM350	Sol. Ann.	160	1103	55	379	40		B95
AM350	DA	195	1344	155	1069	10.5		C41
AM355	Sol. Ann.	175	1207	65	448	30		B95
AM355	DA	195	1344	155	1069	10		C41
Custom 450	Anneal	125	862	95	655	10	40	C30
Custom 450	H900	180	1241	170	1172	10	40	C40
Custom 455	H900	235	1620	220	1517	8	30	C47
Stainless W	Sol. Ann.	120	827	75	517	7		C30
Stainless W	H950	195	1344	180	1241	7	25	C46

Figure 8.2 The mechanical properties of PH stainless steels.

8.3 PHYSICAL PROPERTIES ^{[1] [2] [30] [37]}

8.3.1. General ^[37]

P.H	1,4542 ⁽¹⁾	630	7,8	200	10,8 ⁽⁴⁾	11,6 ⁽³⁾	16	500	0,71	Non magnetic (°)
	1,4568 ⁽²⁾	631	7,8	200	11,0 ⁽⁴⁾	11,6 ⁽³⁾	16	500	0,80	
	1,4980 ⁽³⁾	660	8,0	203 ⁽³⁾	17,5	18,0	13	490	0,91 ⁽⁴⁾	

Figure 8.3 The physical properties of PH stainless steels.

8.3.2. Weldability and Issues ^{[2] [30]}

Methods commonly used with PH stainless steels are:

- Shielded Meta Arc Welding (SMAW)
- Gas Tungsten Arc Welding TIG (GTAW)
- Gas Metal Arc Welding MIG (GMAW)

The martensitic and semi-austenitic PH stainless steels solidify mainly as primary ferrite in either the FA or F mode. The WRC-1992 diagram, however, does not include effects for aluminum or titanium in the chromium equivalent, so an estimate of the effect of these alloying elements was needed to properly portray the solidification of alloys. It was Hull who evaluated these two elements as part of a study of ferrite content in chill cast pins. He found the acceptable chromium-equivalent coefficients of 2.48 for aluminum and 2.20 for titanium. A modified version of the WRC-1992 diagram is shown below:

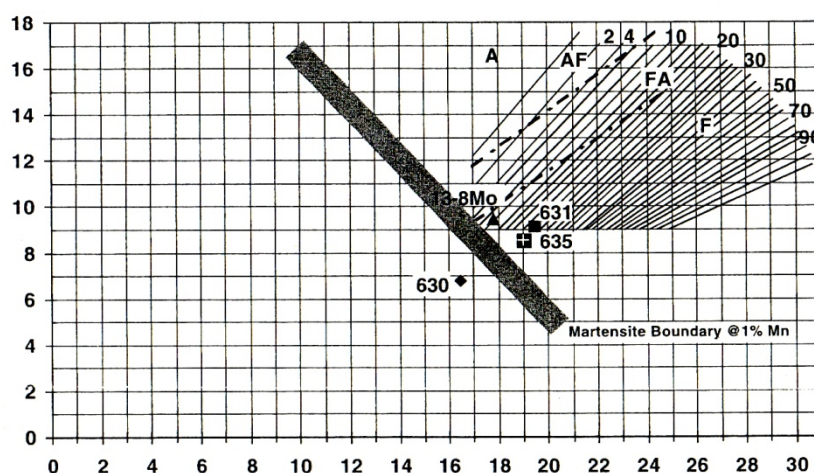


Figure 8.4 The modified version of the WRC-1992 diagram, designed to contain the martensite boundary.

The revised nickel- and chromium-equivalent formulae are shown below:

$$Ni_{eq} = wt\%Ni + 35 \cdot wt\%C + 20 \cdot wt\%N + 0.25 \cdot wt\%Cu$$

$$Cr_{eq} = wt\%Cr + wt\%Mo + 0.7 \cdot wt\%Nb + 2.48 \cdot wt\%Al + 2.2 \cdot wt\%Ti$$

The positions of several martensitic and semi-austenitic stainless steels are plotted on the diagram above. As it can clearly be seen they all solidify as virtually 100% ferrite. The modified WRC-1992 can be used to predict the solidification

conditions, but not the final weld metal microstructure, since much of the austenite will eventually transform to martensite.

When welding PH stainless steels, it is very common to weld them in the solution-treated condition, before precipitation takes place. In this condition the martensitic PH grades are somewhat hard, but still moderately ductile. The semi-austenitic and austenitic PH stainless steels are soft and ductile. Because the weld metal cools rapidly, virtually no precipitation occurs in the weld metal. As a result the weld metal in the as-welded condition is very much like the solution-annealed base metal in terms of its microstructure and properties. However, the weld metal is not as homogenous.

After being welded, the martensitic PH stainless steels are usually given a PWHT in the range of 480 to 620°C (900 to 1150°F). This heat treatment both tempers the martensite and promotes precipitation strengthening. At temperatures over 540°C (1000°F), however, PWHT may cause some austenite to re-form in the structure. Semi-austenitic alloys, containing a high volume of stable austenite, may require a conditioning heat treatment that would cause carbide precipitation and render the austenite less stable. Upon cooling from the conditioning temperature the austenite transforms to martensite. After conditioning at low temperatures in the range of 730 to 760°C (1345 to 1400°F) effectively all the austenite transforms. At higher temperatures in the range of 930 to 955°C (1705 to 1750°F) some austenite may remain after cooling to room temperature. In this case, refrigeration is used to transform the austenite to martensite. The austenitic alloys, on the other hand, are hardened in the temperature range of 700 to 750°C (1260 to 1350°F). Since the austenite is very stable, there is no change in microstructure at these temperatures or on cooling to room temperature.

As long as the PH alloys is in the solution-annealed condition prior to welding and the full hardening treatment is applied, the weld strength obtained is very similar to that of the base metal, even if ductility is somewhat lower. If the base metals are already in the fully hardened condition prior to welding there is a significant risk of cracking. After aging to peak strength, the martensitic and semi-austenitic PH stainless steels have limited ductility, and strains associated with weld shrinkage may be sufficient to produce cracks around the weld. The ductility of the austenitic PH stainless steels in the fully hardened condition is somewhat better than that of the other PH alloys, but still they tend to have much more severe weld cracking problems in general. As such, ***welding of any PH stainless steel in the fully hardened condition is to be avoided.***

In any event, PH stainless steels carry the advantage of being able to strengthen to levels approaching those of high-strength structural steels, while still retaining a remarkably good corrosion resistance. Achieving optimum properties requires careful control of microstructure and heat treatment conditions. When dealing with martensitic and semi-austenitic alloys, excessive levels of retained ferrite can reduce toughness and ductility and as a result ferrite level should be kept below about 10%. The presence of retained austenite, on the other hand, will reduce strength in these alloys. Achieving a good balance of strength and toughness, which is required in many applications, requires heat treatment in the higher temperature range of 550 to 600°C (1020 to 1110°F). In general, however, the peak weld metal strength and the peak base metal strength are quite similar. The weld metal toughness though tends to be somewhat lower than the base metal toughness, except when the weldment is aged to highest strength.

All these heat-treatment conditions prior to welding should be chosen to minimize defects and optimize properties, and also depend on postweld treatments

that will be applied to the welded assembly. The martensitic types are welded in the annealed condition for thin sections and in the overaged condition for thick sections. Thick sections tend to have high restraints and welding them in the overaged condition, which practically puts them in a weakened state, will reduce residual stresses in the welded assembly. Semi-austenitic PH grades are typically welded in the solution-treated or annealed condition. Finally, austenitic PH stainless steels are the most difficult types to weld because of **solidification**, **liquation** and **ductility-dip cracking** problems. These are normally welded in the solution-annealed condition, to minimize the risk of these dangers. TiC is a particle that has been identified of being subject to constitutional liquation; a phenomenon primarily responsible for liquation cracking in the HAZ and/or PMZ. Minimizing grain size and using low heat inputs will allow for steep temperature gradients that would actually minimize the amount and extent of liquation that occurs in situations such as this.

On another note, **weld solidification cracking** is a persistent problem among austenitic PH alloys. The fusion zone of these alloys is fully austenitic with no delta ferrite and adjusting the composition to promote FA-type solidification is actually not an option. Reducing impurity elements, particularly sulphur, phosphorous and silicon (all elements which produce low-melting-point compounds) seems to be the only way to prevent this type of cracking.

Finally, **hydrogen-induced cracking** does not seem to be a problem for the PH stainless steels, mainly due to their low carbon content and resulting low hardness. When dealing with thick sections of martensitic and semi-austenitic PH stainless steels, however, low-hydrogen practices are recommended because these PH grades may be transformed to martensite before any PWHT occurs.

8.3.3. Corrosion and Erosion Resistance ^{[2] [25] [30] [47]}

Corrosion resistance of PH stainless steels appears to depend on the condition of heat treatment that is applied on the specific alloy. In general, the corrosion resistance is optimized when the steel is in the fully hardened condition, but this cannot be used as a rule of thumb since there are exceptions for specific corrosive media. Consequently, it is best to consult the recommendations of the steel supplier for specific alloys. For example, the 17-4PH alloy is claimed to show welded corrosion resistance comparable to the unwelded condition, when the aging heat treatment is applied after welding. The 17-7PH alloy, on the other hand, exhibits corrosion resistance comparable to that of Type 304 stainless steel in most media and significantly superior to that of the hardenable 400 series stainless steels.

WELDING IN SHIPBUILDING

9.1 A GENERAL WALKTHROUGH ^[55] ^[56]

Welding is a process that joins materials using coalescence. This can either be done by melting the workpieces and adding a filler material to form a pool of molten material that cools to become a strong joint, or by melting a lower-melting-point material between the workpieces to form a bond between them. There are many different welding processes, many types of welds, and many ways to make a weld. Some welding processes do not cause sparks, use electricity, or require added heat. Many different energy sources can be used for welding including a gas flame, an electric arc, a laser, an electric beam, friction, or even ultrasound.

The result of this diversity is a vast variety of welding processes the modern engineer can choose from today. In order to make the right choice one should first be aware of the process best suited for the materials at hand. To that extent, the following table is a list of welding processes used to weld various metals, followed by a rating indicating the processes' compatibility to each metal:

Base Metals Welded	Welding Processes								
	SMAW	GTAW	PAW	SAW	GMAW	FCAW	ESW	BW	OFW
Aluminums	C	A	A	No	A	No	Exp	B	B
Copper-base alloys									
Brasses	No	C	C	No	C	No	No	A	A
Bronzes	A	A	B	No	A	No	No	A	B
Copper	C	A	A	No	A	No	No	A	A
Copper nickel	B	A	A	No	A	No	No	A	A
Irons									
Cast, malleable, nodular iron	A	B	B	No	B	B	No	A	A
Wrought iron	A	B	B	A	A	A	No	A	A
Lead	No	B	B	No	No	No	No	No	A
Magnesium	No	A	B	No	A	No	No	No	No

(table continued next page)

Base Metals Welded	Welding Processes								
	SMAW	GTAW	PAW	SAW	GMAW	FCAW	ESW	BW	OFW
Nickel-base alloys									
Inconel	A	A	A	No	A	No	No	A	B
Monel	A	A	A	C	A	No	No	A	A
Nickel	A	A	A	C	A	No	No	A	A
Nickel silver	No	C	C	No	C	No	No	A	B
Precious metals	No	A	A	No	Exp	No	No	A	B
Steels									
Alloy steel	A	A	A	B	A	A	A	A	A
Low-alloy steel	A	A	A	A	A	A	A	A	A
High- and medium-carbon steel	A	A	A	B	A	A	A	A	A
Low-carbon steel	A	A	A	A	A	A	A	A	A
Stainless steel	A	A	A	A	A	B	A	A	C
Tool steel	A	A	A	No	C	No	No	A	A
Titanium	No	A	A	Exp	A	No	No	No	No
Tungsten	No	B	A	No	No	No	No	No	No
Zinc	No	C	C	No	No	No	No	No	C
A: recommended or easily weldable B: acceptable or weldable with precautions C: possibly usable but not popular or restricted use or difficult to weld No: not recommended or not weldable Exp: experimental or research									

Welding process selection, however, depends not only on the materials being welded, but also on the service application the weldment is intended for. Following is a detailed list of welding processes, separated into categories depending on the way they function. The table's right column lists the applications where each process is used:

ARC WELDING		
Name	Characteristics	Applications
Atomic hydrogen welding	Two metal electrodes in hydrogen atmosphere	Historical
Bare metal arc welding	Consumable electrode, no flux or shielding gas	Historical
Carbon arc welding	Carbon electrode, historical	Copper, repair (limited)
Electrode gas welding	Continuous consumable electrode, vertical positioning, steel only	Storage tanks, shipbuilding
Electroslag welding	Continuous consumable electrode, vertical positioning, steel only	Storage tanks, shipbuilding
Flux cored arc welding	Continuous consumable electrode filled with flux	Industry, construction
Gas metal arc welding ¹	Continuous consumable electrode and shielding gas	Industry
Gas tungsten arc welding ²	Non-consumable electrode, slow, high quality welds	Aerospace
Plasma arc welding	Nonconsumable electrode, constricted arc	Tubing, instrumentation
Shielded metal arc welding ³	Consumable electrode covered in flux, can weld any metal as long as they have the right electrode	Construction, outdoors

(table continued next page)

ARC WELDING		
Name	Characteristics	Applications
Stud arc welding	Welds studs to base material with heat and pressure	Construction, <i>shipbuilding</i>
Submerged arc welding	Automatic, arc submerged in granular flux	<i>Ship industry</i>
1. Also known as metal inert gas (MIG) or metal active gas (MAG) welding. 2. Also known as tungsten inert gas (TIG) welding. 3. Also known as manual metal arc (MMA) welding or stick welding.		

OXYFUEL GAS WELDING		
Name	Characteristics	Applications
Air acetylene welding	Chemical welding process, not popular	Limited
Oxyacetylene welding	Combustion of acetylene with oxygen produces high-temperature flame, inexpensive equipment	Maintenance, repair
Oxyhydrogen welding	Combustion of hydrogen with oxygen produces flame	Limited
Pressure gas welding	Gas flames heat surfaces and pressure produces the weld	Pipe, railroad rails (limited)

RESISTANCE WELDING		
Name	Characteristics	Applications
Flash welding		
Pressure-controlled resistance welding		
Projection welding		
Resistance seam welding	Two wheel-shaped electrodes roll along workpieces, applying pressure and current	
Resistance spot welding	Two pointed electrodes apply pressure and current to two or more thin workpieces	Automobile industry
Upset welding	Butt joint surfaces heated and brought together by force	

SOLID-STATE WELDING		
Name	Characteristics	Applications
Coextrusion welding		
Cold welding		
Diffusion welding		
Explosion welding	Military hardware.	
Forge welding		
Friction welding	Thin heat affected zone, need sufficient pressure	Aerospace industry, railway, land transport
Friction stir welding		
Hot pressure welding		
Roll welding		<i>Dissimilar materials</i>
Ultrasonic welding		

OTHER WELDING METHODS		
Name	Characteristics	Applications
Electron beam welding	Deep penetration, fast, high equipment cost	
Electroslag welding	Welds thick workpieces quickly, vertical position, steel only	Heavy plate fabrication, construction
Flow welding		
Induction welding		
Laser beam welding	Deep penetration, fast, high equipment cost	Automotive industry
Laser-hybrid welding	Combines LBW with GMAW in the same welding head, able to bridge gaps up to 2mm (between plates), previously not possible with LBW alone.	Automotive, Shipbuilding , Steelwork industries
Percussion welding		
Thermite welding		Railway tracks

As one can clearly see, many of these welding processes can be used to join stainless steels together or to other dissimilar metals, however there are only a few of them that are actually used in shipbuilding. It is these welding processes that are more commonly used in shipbuilding and in the ship-industry in general that will be described in detail below.

9.2 STUD WELDING^{[55] [56] [57]}

Arc stud welding (SW) is an arc welding process that uses an arc between a metal stud, or similar part, and the other workpiece. The process is used without filler metal, with or without shielding gas or flux, with or without partial shielding from a ceramic ferrule surrounding the stud, and with the application of pressure after the faying surfaces are sufficiently heated. The following figure shows the schematic representation of a stud weld gun and its power source.

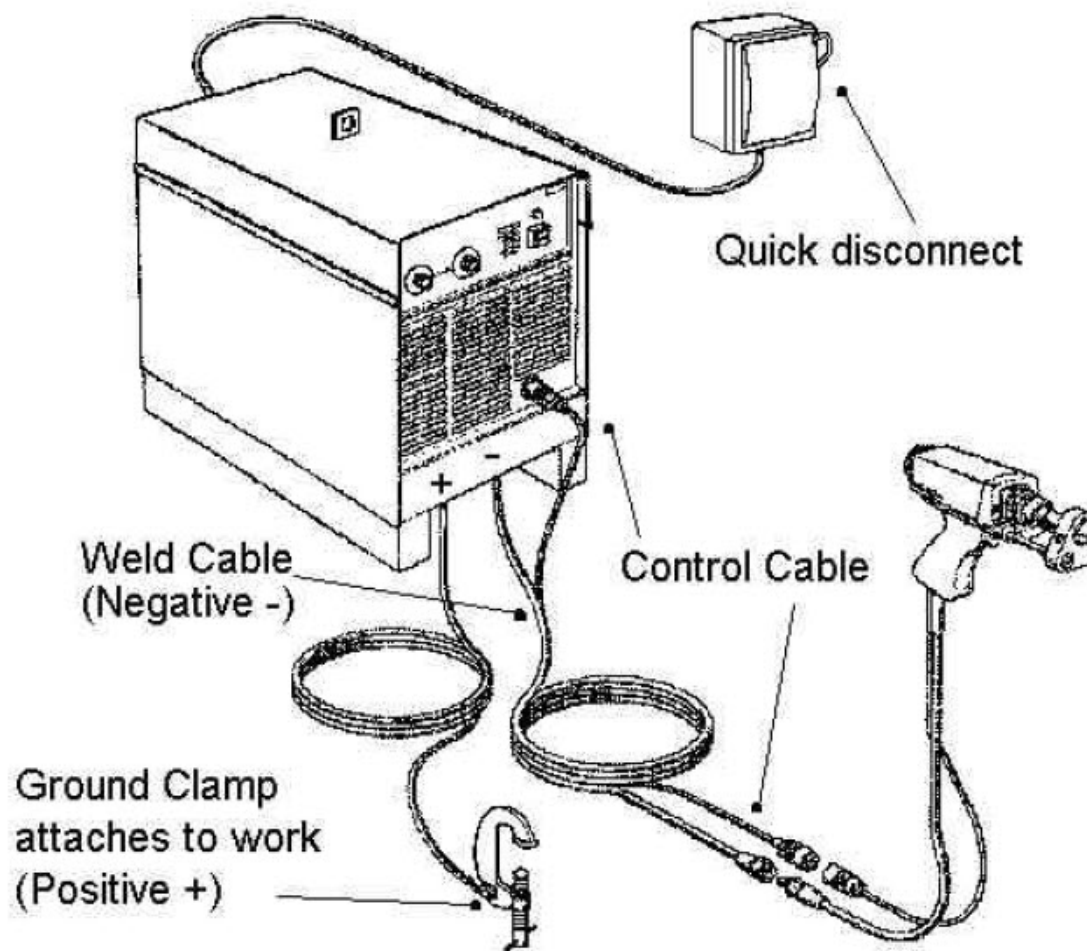


Figure 9.1 Stud weld apparatus.

Stud welding has a very specialized field of application and is not a metal joining process in the same manner as other welding processes. This process is actually a combination of arc welding and forge welding and thus it is questionable whether it is a true arc welding process. It is based on two steps. First, electrical contact between the end of the stud and base metal occurs and an arc is established. The heat of the arc melts the surface of stud's end and the work surface. As soon as the entire cross section of the stud and an area of equal size on the base metal are melted, the stud is forced against the base metal. The molten end of the stud joins with the molten pool on the work surface, and as the metal solidifies the weld is produced. Partial shielding is normally accomplished by means of a ceramic ferrule that surrounds the arc area and by fluxing ingredients sometimes placed on the arcing end of the stud. The following figure shows how a stud weld is made.

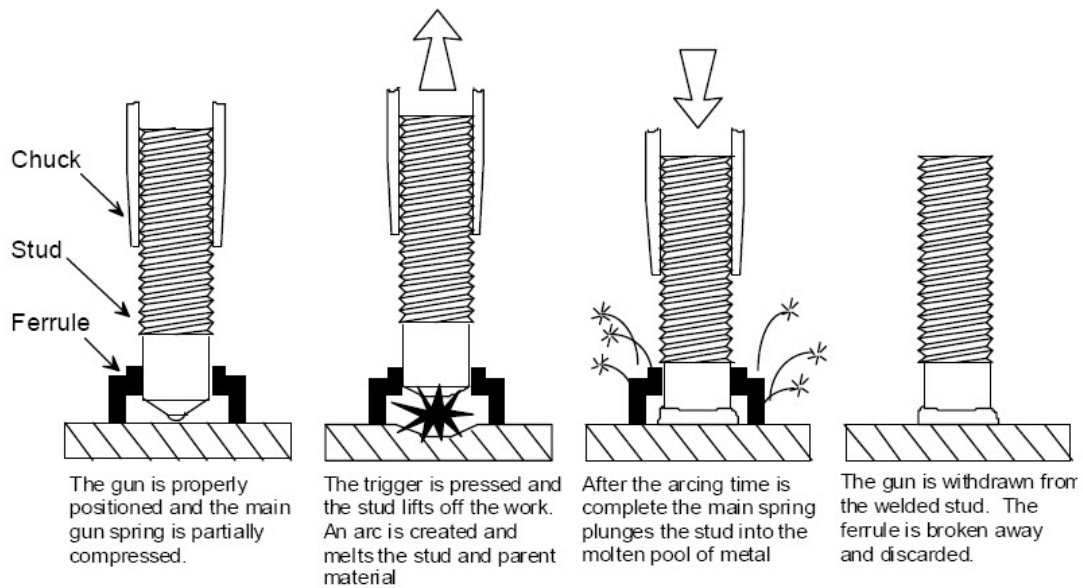
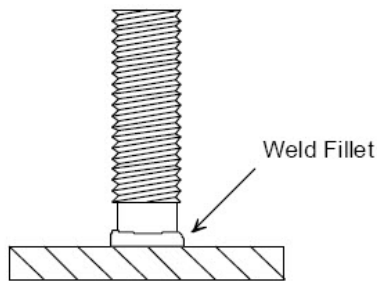


Figure 9.2 The making of a stud weld.

The arc stud welding process offers tremendous cost savings when compared to drilling and tapping for studs or to manually welding studs to base metal. It also does not compromise water tightness or weaken the base metal. This process is widely used for welding studs to mild steels, low-alloy steels, and some of the austenitic types of all stainless steels only, as long as the stud has the same analysis as the base metal. The minimum recommended plate thickness to permit efficient welding without burn-through or excessive distortion is expressed as a general rule. The general rule states that the minimum thickness of the plate or base metal is 20% of the stud base diameter. To develop full strength of the stud, the plate thickness should not be less than 33% of the stud base diameter.

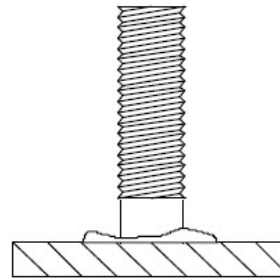
The semiautomatic method of applying studs is the most common for construction work and especially for ship work. In this method of applying the welder holds the stud gun during the welding operation. When the trigger is pressed, the entire weld cycle is automatic, based on the settings of the control unit. The gun should not be moved during the welding operation. The arc will be noticed, and the plunging action will be heard at the completion of the welding cycle. After the plunging operation, the gun should be held steady for at least a half-second before withdrawing it from the welded stud.

A variety of problems might occur in making a stud weld. If the current is too low, a full-strength weld will not result. If it is too high, there will be too much metal discharged from the weld. The time cycle must also be set correctly. If the plunge is too short, a full weld will not result. If the ferrule is not in place properly, the weld will be off-side. If the gun is not perpendicular to the workpiece, the stud will be welded at an angle. As a result, every time a new job is started or a new setup is made, the welding procedure should be verified by bending a stud with hammer blows to see that the weld procedure produces a quality weld. The following figure represents some of the most common defects encountered in arc stud welding.



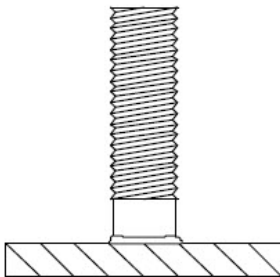
GOOD WELD

A **good weld** is characterized by a clean even fillet of good height all the way (360°) around the weld stud. The fillet should be bright and shiny with a bluish color and have a slight flow of metal into the base material from the bottom to the fillet. Another good indication of weld integrity is to check the burn-off. Place a stud next to one that has been welded, the welded one should be approximately 1/8" shorter.



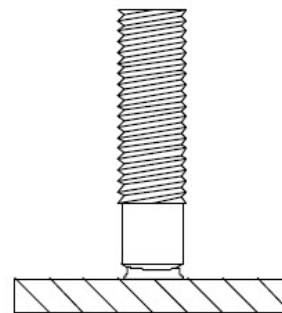
HOT WELD

A **hot weld** can be identified by a concave weld fillet that is close to the work surface (washed out fillet). Additionally, there may be excessive weld material spattered beyond the area contained by the ferrule. You may also be able to notice burn through on the base material. Corrective action is to decrease time, power or both.



COLD WELD

A **cold weld** is indicated if the weld fillet is not formed completely around the weld stud (360°) or there is low fillet height this may indicate a cold weld. *NOTE: This may also be an indication of too little plunge. Verify the plunge setting first!* Another sign of a cold weld is a dull grayish appearance and/or stringers ("spider legs") coming from the weld fillet. Corrective action is to increase time, power or both.



HANG UP

A **hang up** occurs when the stud base is partially melted away and the stud appears to be perched on a small portion of its base. This is caused by a bind in the gun that prevents the stud from plunging smoothly. A bind in the gun shaft will cause this, but most often it is caused by the stud rubbing the ferrule. For a good weld to occur consistently, it is extremely important that the stud be exactly centered in the ferrule. This is accomplished by adjusting the gun foot.

Figure 9.3 Examples of various stud welds.

In the shipbuilding industry arc stud welding is used for attaching wood decking to metal decking, as well as for attaching insulation to the interior steel portions of ships.

9.3 SUBMERGED ARC WELDING ^{[55] [56] [58]}

Submerged arc welding (SAW) is an arc welding process that uses an arc or arcs between a bare metal electrode or electrodes and the weld pool. A blanket of granular flux on the workpieces shields the arc and molten metal. The process is used without pressure and with filler metal from the electrode and sometimes from a supplemental source (welding rod, flux, or metal granules). This normally applied automatic welding process was also known as *under powder welding* or *smothered arc welding*.

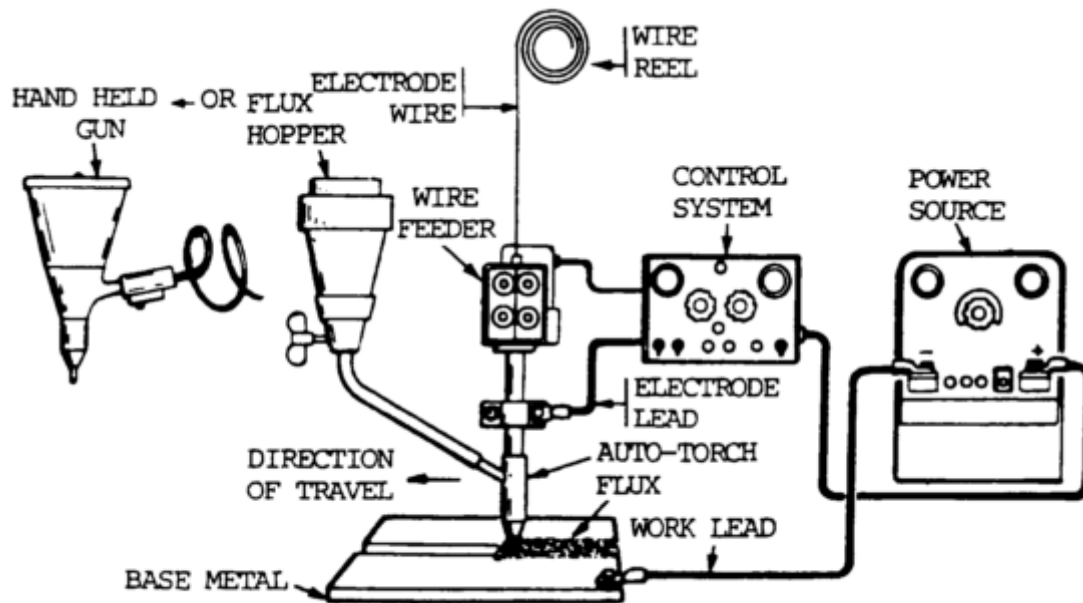


Figure 9.4 A submerged arc welding apparatus.

The SAW process uses the heat of an arc between a continuously fed electrode and the workpiece. The heat of the arc melts the surface of the base metal and the end of the electrode. The metal melted off the electrode is transferred through the arc to the workpiece, where it becomes the deposited weld metal. Shielding is obtained from a blanket of granular flux, which is laid directly over the weld area. The flux close to the arc melts and intermixes with the molten weld metal, helping purify and fortify it. The flux forms a glass-like slag that is lighter in weight than the deposited weld metal and floats to the surface as a protective cover. The weld is submerged under this layer of flux and slag, hence the name of the process. The flux and slag normally cover the arc so that it is not visible. The unmelted portion of the flux can be reused. The electrode is fed automatically from a coil into the arc, which is also maintained and initiated by a fuse-type start.

The SAW process is a limited position welding process. Welding can be done in the flat position and in the horizontal fillet position with ease. The welding positions are limited because of the large pool of molten metal which is very fluid. The slag is also fluid and will tend to run out of the joint. Under special controlled procedures it is possible to weld in the horizontal position, sometimes called 3-o'clock welding. This, however, requires special devices to hold the flux in place, so that the molten slag and weld metal cannot run away. The process is not used in the vertical or overhead position. The positioning of the electrode also has significant effect on the weld, as shown on the following figure.

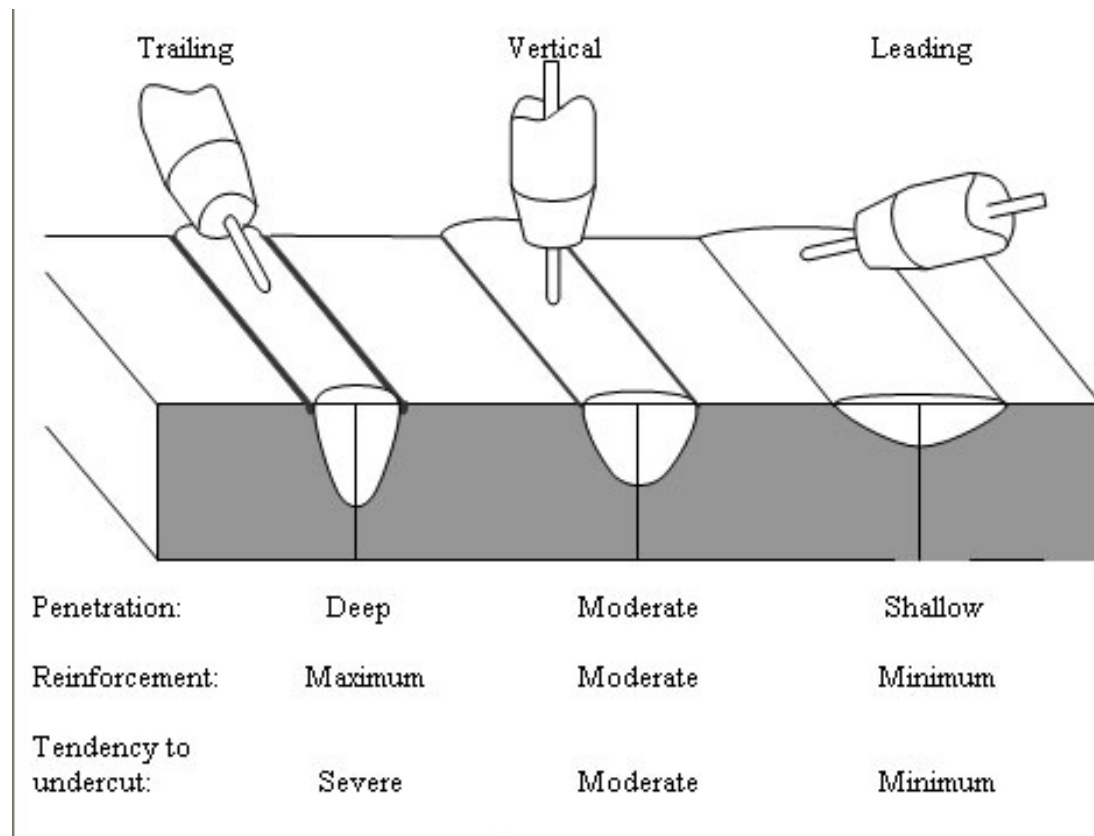


Figure 9.5 Results for various filler metal orientation.

SAW is used to weld low- and medium carbon steels, HSLA steels, quenched and tempered steels, and of course many stainless steels. Experimentally it has been used to weld certain copper alloys, nickel alloys, and even uranium.

The SAW process was originally used to make the longitudinal seam in large pipes. Some of its major advantages are the high-quality weld metal it can provide and its extremely high deposition rate and speed that surpass the deposition rate and speed of any other arc welding process. It produces a smooth, uniform finished weld with no spatter and little or no smoke at all. SAW is widely used in the shipbuilding industry for splicing and fabricating subassemblies, but also by many other industries where steels are used in medium to heavy thickness. The SAW process also exhibits higher heat input and slower cooling rates than most arc welding processes, thus allowing gases more time to escape and creating less defects in the weld metal.

The submerged arc welding process carries only one major drawback; its limited welding positions capability. SAW can only be used when the axis of the weld is vertical. The only other limitation attributed to this process is the fact that it is used primarily to weld steels, since its high-heat input and slow cooling cycle can be an issue when welding quenched and tempered steels. However, since this thesis is about the welding of stainless steels in shipbuilding, this limitation is of no concern.

Nonetheless, several problems may occur when using the semiautomatic application method. The electrode wire may be curved when it leaves the nozzle of the welding gun. This curvature can cause the arc to be struck in a location not expected by the welder. When welding in deep grooves, the curvature may cause the arc to be against one side of the weld joint, rather than at the root. This will cause incomplete root fusion, and the flux will be trapped at the root of the weld.

Another problem with semiautomatic welding is the problem of completely filling the weld groove or maintaining exact size, since the weld is hidden and cannot be observed while it is being made. This requires making an extra pass, where in some cases too much weld is deposited. Another quality problem is associated with the extremely large single-pass weld deposits. When these large welds solidify, the impurities in the melted base metal and in the weld metal all collect at the last point to freeze, which is the centerline of the weld. If there is sufficient restraint and enough impurities are collected at this point, centerline cracking may occur. This can actually happen when making large single-pass flat fillet welds if the base metal plates are 45° from flat. A simple solution is to avoid placing the parts at a true 45° angle. It should be varied approximately 10° so that the root of the joint is not in line with the centerline of the fillet weld. Another solution is to make multiple passes, rather than attempting to make a large weld in a single pass. In automatic and mechanized welding, defects may occur at the start or at the end of the weld. The best solution remains the use of runout tabs so that starts and stops will be on the tabs, rather than on the product.

A most interesting variation of the SAW process is the **SAW strip cladding** process, which is used primarily to overlay mild and alloy steels, usually with stainless steel. A strip electrode is used as the (usually positive) electrode and an electric arc is formed between the strip and the workpiece. Flux is used to form a molten slag to protect the weld pool from the atmosphere and helps to form a smooth weld bead surface. The bead has a uniform and minimum penetration. It is used for overlaying the inside of vessels to provide the corrosion resistance of stainless steel, while still using the strength and economy of the low-alloy steels for the wall thickness. When the width of the strip is over 50 mm (2 in.), a magnetic arc oscillating device is employed to provide for even burn-off of the strip and uniform penetration.

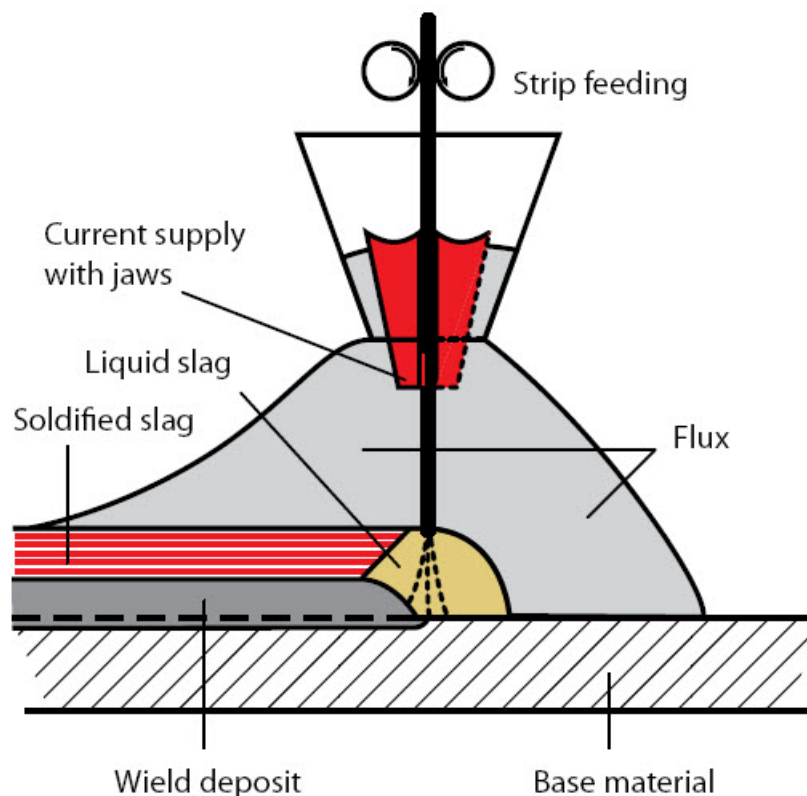


Figure 9.6 Schematic representation of a SAW strip cladding.

9.4 ELECTROSLAG WELDING ^[55] ^[56] ^[58]

Electroslag welding (ESW) is a welding process that produces coalescence of metals with molten slag that melts the filler metal and the surfaces of the workpieces. The weld pool is shielded by the slag, which moves along the full cross section of the joint as welding progresses. The process is initiated by an arc that heats the slag, which is then kept molten by its resistance to electric current passing between the electrode and the workpieces. Normally, molding shoes are used to confine the molten weld metal and slag; pressure is not used. This welding process is normally used to make welds in the vertical position and on steels and has two major variations: the consumable guide system and the non-consumable guide system, as they are both respectively shown on the following figures. Electroslag welding is a well-established method for joining thick components for bridges, buildings, ships, pressure vessels, and more.

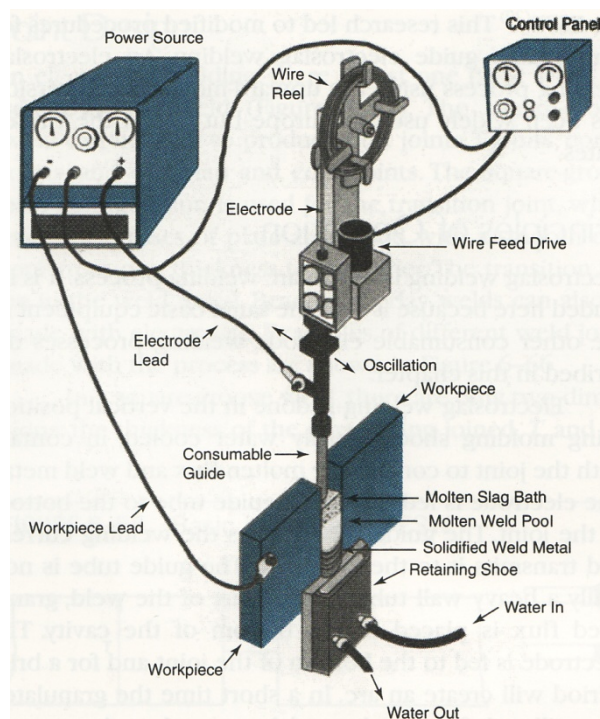


Figure 9.7 An electroslag welding apparatus.

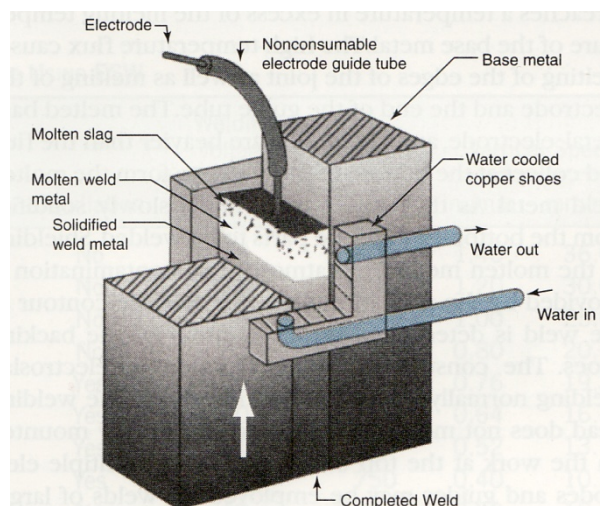


Figure 9.8 A schematic representation of an electroslog welding.

Electroslog welding is not an arc welding process *per se*. It is included in the arc welding processes because it uses the same basic equipment as the other consumable electrode welding processes. It is done in the vertical position using molding shoes, usually water cooled, in contact with the joint to contain the molten flux and weld metal. The electrode is fed through a guide tube to the bottom of the joint. The guide tube carries the welding current and transmits it to the electrode. This guide tube is normally a heavy wall tube. At the start of the weld, granulated flux is placed in the bottom of the cavity. The electrode is fed to the bottom of the joint and for a brief period will create an arc. In a short time the flux will melt from the heat of the arc and produce a pool of molten flux. The flux is electrically conductive and the welding current will pass from the electrode through the molten flux to the base metal. The passage of current through the conductive flux causes it to become hot, and it reaches a temperature in excess of the melting temperature of the base metal. The high-temperature flux causes melting of the edges of the joint as well as melting of the electrode and the end of the guide tube. The melted base metal, electrode, and guide tube are heavier than the flux and collect at the bottom of the cavity to form the molten weld metal. As the molten weld metal slowly solidifies from the bottom, it joins the parts to be welded. Shielding of the molten metal from atmospheric contamination is provided by the pool of molten flux. Surface contour of the weld is determined by the contour of the backing shoes. Multiple electrodes and guides may be employed for welds of larger cross section or it is also possible to oscillate the electrode and guide tube across the width of the joint. The surface of the solidified weld metal is covered with an easily removable thin layer of slug. The slag loss must be compensated for by adding flux during the welding operation. A starting tab is necessary to build up the proper depth of the slag so that the molten pool is formed at the bottom of the joint. Runoff tabs are also required at the top of the joint so that the molten flux will rise above the top of the joint. Both starting and runoff tabs are removed from the ends of the joint once the weld is completed.

The metals welded by this method are low-carbon steels, HSLA steels, medium-carbon steels, and certain stainless steels. Quenched and tempered steels can be electroslog welded; however, a PWHT is necessary to compensate for the softened heat-affected zone.

Electroslog welding is a low-hydrogen welding process; hydrogen is not present in any of the materials involved in making the weld, evidently making it a perfect process for producing welds that exhibit no risk for hydrogen-induced cracking. Because of the slow cooling rate, any impurities that are in the base metal and are melted during the welding process have enough time to escape. This slow cooling rate allows large grain growth in the weld metal and the HAZ and also minimizes the risk of cracking and reduces the hardness in the HAZ sometimes found in conventional arc welds. Weld metal produced by this method will qualify under the most strict codes and specifications.

The only major limitation of this process is its welding position. Electroslog welding can be used only when the axis of the weld is vertical. A tilt of up to 15° is permitted, but beyond this the process may not function correctly. Its second limitation of only being usable on steels is not an issue for this thesis.

A variation of the electroslog welding, which particularly applies to shipbuilding, is **electroslog cladding**. This is a variation that deposits surfacing materials on base metals. It is very similar to **strip cladding** with the **submerged arc welding** process except that the heat required to melt the surface of the base metal, the strip, and

the flux is generated by resistance heating from the current flow to the strip and through a shallow layer of electroconductive slag. There is no arc between the strip electrode and the parent material. The following figure is a schematic representation of an ESW strip cladding equipment.

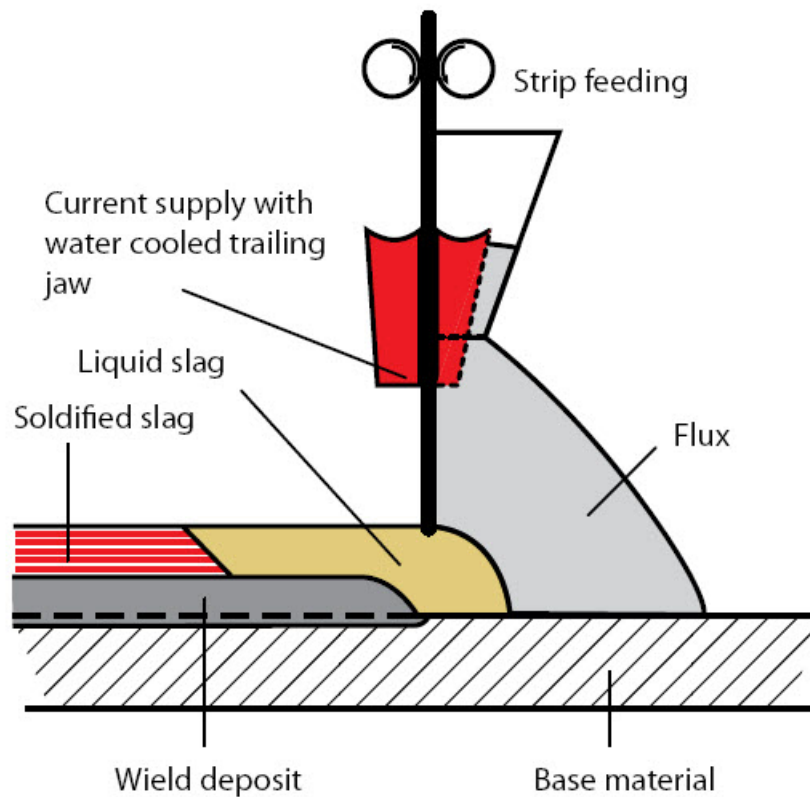


Figure 9.9 A schematic representation of an electroslog strip cladding.

The penetration achieved with ESW strip cladding is less than that with SAW strip cladding because it is the molten slag pool that is used to melt the strip and a portion of the parent material. The temperature of the slag pool is about 2300°C (4200°F), making it necessary to water-cool the contact jaws. Compared to SAW strip cladding, ESW strip cladding has an increased deposition rate of 60% to 80% and only half of the dilution (10% - 15%) from the base material due to less penetration. However, it can be used with the same equipment as that used for SAW strip cladding. It also has an increased welding speed (50% - 200%) and a lower flux consumption (about 0.5 kg/kg strip). Its most important trait, however, is that the solidification rate of the ESW clad weld metal is lower than that of the SAW clad weld metal, thus aiding the de-gassing process and increasing the resistance to porosity. Oxides can rise easier out the molten pool to the surface resulting in a metallurgically cleaner weld metal, which is less sensitive to hot cracking and corrosion.

Using this method it is possible to clad ferritic, martensitic, and austenitic stainless steels, nickel-base alloys, and some hard-surfacing materials. Nonetheless, the major use for ESW cladding is the deposition of austenitic stainless steel on carbon steel for tube sheets and other corrosive-resistant applications or depositing corrosion-resistant coatings on large shafts.

At this point, having described both SAW and ESW strip cladding processes, it is considered useful to present the following table, which contains a collection of all possible or recommended combinations for SAW and ESW strip cladding.

		Nr.		Layer 1 ⁽²⁾	Layer 2	A	V	cm/min
Low alloy	SAW	1	OK Flux 10.31	OK Band 7018		750	28	12
	SAW	2	OK Flux 10.31	OK Band 7018	OK Band 7018	750	28	12
306 L	SAW	2	OK Flux 10.05	OK Band 309L	OK Band 308L	750	28	13
	ESW	1	OK Flux 10.10	OK Band 309L ESW	-	1250	24	18
	ESW	2	OK Flux 10.10	OK Band 309L ESW	OK Band 309L ESW	1250	24	18
	SAW/ESW ⁽³⁾	2	OK Flux 10.05/10.10	OK Band 309L	OK Band 308L	1250	24	32
316 L	SAW/ESW ⁽³⁾	2	OK Flux 10.05/10.14	OK Band 309L	OK Band 308L	1250	24	32
	SAW	2	OK Flux 10.05	OK Band 309L	OK Band 316L	750	28	13
	SAW	1	OK Flux 10.06	OK Band 309L ⁽⁴⁾	-	1125	27	26
	SAW	1	OK Flux 10.06F	OK Band 309L	-	750	28	12
	ESW	1	OK Flux 10.10	OK Band 309LMo ESW	-	1250	25	18
	ESW	2	OK Flux 10.10	OK Band 309LMo ESW	OK Band 309LMo ESW	1250	25	18
	SAW/ESW ⁽⁵⁾	2	OK Flux 10.05/10.10	OK Band 309L	OK Band 316L	1250	24	32
	SAW/ESW ⁽⁵⁾	2	OK Flux 10.05/10.14	OK Band 309L	OK Band 316L	2000	26	35
317L	SAW	2	OK Flux 10.05	OK Band 309L	OK Band 317L	750	28	12
347	SAW	2	OK Flux 10.05	OK Band 309L	OK Band 347	750	28	13
	SAW	1	OK Flux 10.05	OK Band 309LNb	-	750	28	12
	ESW	1	OK Flux 10.10	OK Band 309LNb ESW	-	1250	25	18
	ESW	2	OK Flux 10.10	OK Band 309LNb ESW	OK Band 309LNb ESW	1250	24	18
	ESW	1	OK Flux 10.14	OK Band 309LNb	-	2300	24	40
	ESW	1	OK Flux 10.14	OK Band 309LNb ⁽⁴⁾	-	2300	24	30
	SAW/ESW ⁽⁵⁾	2	OK Flux 10.05/10.10	OK Band 309L	OK Band 347	1250	24	18
	SAW/ESW ⁽⁵⁾	2	OK Flux 10.05/10.14	OK Band 309L	OK Band 347	2000	26	35
2209	SAW	2	OK Flux 10.05	OK Band 2209	OK Band 2209	750	28	12
904L	SAW	3	OK Flux 10.05	OK Band 385	OK Band 385	750	28	12
	ESW	1	OK Flux 10.11	OK Band 385		1250	24	18
	ESW	2	OK Flux 10.11	OK Band 385	OK Band 385	1250	24	18
310 MoL	ESW		OK Flux 10.10	OK Band 310MoL	OK Band 310MoL	1250	25	18
410 NiMo	SAW	3	OK Flux 10.07	OK Band 430	OK Band 430 ⁽⁶⁾	770	25	22
430	SAW	2	OK Flux 10.03	OK Band 430	OK Band 430	750	28	12
Alloy 82	SAW	2	OK Flux 10.16	OK Band NiCr3	OK Band NiCr3	750	28	12
	ESW	2	OK Flux 10.11	OK Band NiCr3	OK Band NiCr3	1200	24	25
Alloy 625	SAW	2	OK Flux 10.16	OK Band NiCrMo3	OK Band NiCrMo3	750	27	13
	SAW	3	OK Flux 10.16	OK Band NiCrMo3	OK Band NiCrMo3 ⁽⁶⁾	750	27	13
	ESW	2	OK Flux 10.11	OK Band NiCrMo3	OK Band NiCrMo3	1200	24	25
Monel	SAW	2	OK Flux 10.18	OK Band NiCu7	OK Band NiCu7	750	29	14
	SAW	3	OK Flux 10.18	OK Band NiCu7	OK Band NiCu7 ⁽⁶⁾	750	29	14

1. Strip dimension 60x0.5, if no other information is given
2. Buffer layer if more than one layer is welded
3. Results for second layer (Buffer layer cladded by SAW 750A, 28V, 20 cm/min)
4. Strip dimension 90x0.5 mm
5. Results for second layer (Buffer layer cladded by SAW 750A, 28V, 14 cm/min)
6. Second and third layer
7. For each layer

9.5 ELECTROGAS WELDING ^{[55] [56] [59]}

Electrogas welding (EGW) is an arc welding process that uses an arc between a continuous filler metal electrode and the weld pool, employing vertical position welding with backing to confine the molten weld metal. The process is used with or without an externally supplied shielding gas and without the application of pressure. It is a limited-position arc welding process and is a single-pass process that produces square-groove welds for butt and tee joints. There are two basic variations of the process: one uses the upward-moving-head variation, the other variation uses the consumable guide tube system for use with the electroslag welding. In addition, both of these variations have another two variations. One uses the solid consumable electrode wire and externally supplied shielding gas, normally CO₂. The second uses a flux-cored electrode wire and does not use an external shielding gas, since shielding gases produced by the flux-cored electrode as it is consumed in the arc.

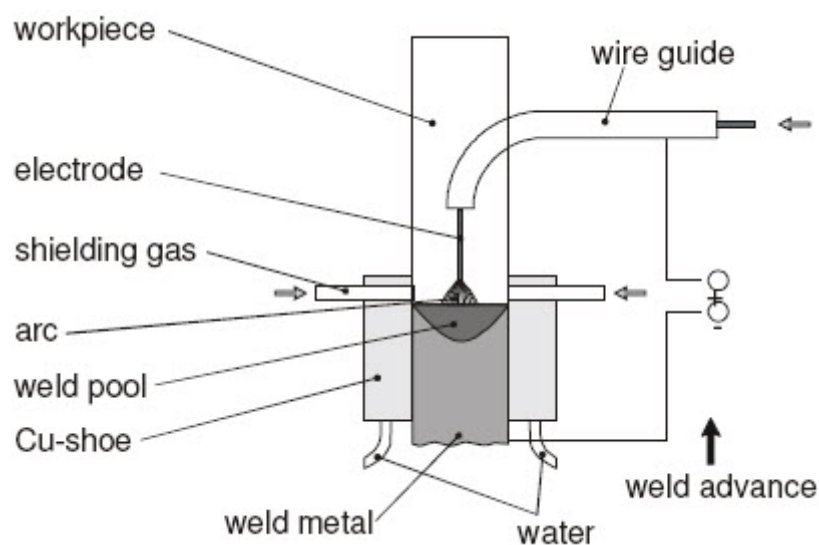


Figure 9.10 An electrogas welding apparatus and a schematic representation of the process.

The primary difference between electrogas welding and electroslag welding is that the arc is continuous in the electrogas process. EGW uses an arc between a continuously fed consumable electrode wire and the molten weld pool. The heat of the arc melts the surface of the base metal and the end of the electrode. The metal melted from the electrode, the metal melted from the surface of the abutting joints, and the metal melted from the consumable guide tube (when used) collect at the bottom of the cavity formed between the parts and the molding shoes. This molten weld solidifies from the bottom of the joint and joins the part to be welded. Shielding of the molten metal from the atmosphere is provided by a shielding gas from the external source or from the disintegration of the ingredients in the cored electrode wire. In the moving-head version, the electrode is fed to the bottom of the joint by means of the wire feeder guide tube or contact tip. This assembly will travel vertically along the joint to maintain the normal arc length between the electrode and the molten weld metal. In some cases one backing shoe is stationary and can be made of steel, thus becoming a part of the joint. Otherwise, the molding shoe can be made of copper and does not become part of the joint, but moves upward. On the side with the wire feeding mechanism, the moving shoe is normally employed, which rises with the wire feeder assembly to maintain the weld metal within the cavity. Normally, only one electrode wire is used for making a weld. In the consumable

guide variation, external shielding gas is not used, but the consumable electrode wire must be of the flux-filled type. Both shielding gas and flux are fed into the arc. The flux contained within the electrode wire turns into slag, which covers the weld deposit. Since the correct amount of flux cannot be provided based on the plate thickness and the root opening, a surplus of flux will be generated. The flux must not accumulate and become too deep, extinguishing the arc. If so the operation would become an electroslog weld. The excess molten flux leaks through the slots in the water-cooled retaining shoes to avoid flux buildup.

The metals welded by the EGW process are low-carbon steels, HSLA steels, medium-carbon steels, and certain stainless steels. The process can also be used for welding quenched and tempered steels, providing that the correct heat input is maintained for the type of steel being welded.

Electrogas welding is considered a low-hydrogen type of welding process, since hydrogen is not present in any of the materials involved in making a weld. Electrogas welds possess properties and characteristics that surpass welds made with shielded metal arc welding processes. The higher-than-normal heat input of electrogas welds reduces the cooling rate, which then helps reduce impurities. This allows larger grain growth of the weld metal and in the HAZ of the base metal. This lower cooling rate minimized the risk of cracking and reduces the high-hardness zone sometimes found in the weld metals and HAZ's produced by shielded metal arc welding processes. The hardness of the weld in EGW is normally uniform across the weld's cross section and is very similar to the unaffected base metal. Weld metal produced by EGW will qualify under most codes and specifications. The ductility of the weld metal of EGW is relatively high, in the range of 25% elongation.

The electrogas welding process is a limited position process. This is also the only major limitation in this welding process. It can be used only when the axis of the weld joint is vertical or varies from the vertical position by no more than 15°. The other limitation of only being usable for welding steels is not an issue when concerned in welding stainless steels. The major use of EGW has been in the field erection of storage tanks and in the shipbuilding industry, for joining shell plates.

9.6 LASER-HYBRID WELDING ^{[55] [56] [60] [61] [62]}

Hybrid laser-arc welding (HLAW or laser-hybrid welding) is a welding process that combines the best attributes of laser welding with those of conventional gas metal arc welding (GMAW). Although the combination of laser light and an electrical arc into an amalgamated welding process has been known since the 1970's, it has only recently been used in industrial applications. There are three main types of hybrid welding process, depending on the arc used; TIG, Plasma arc or MIG augmented laser welding. While TIG augmented laser welding was the first to be researched, MIG is the first to go into industry and is commonly known as hybrid-laser welding. HLAW produces deep penetration welds comparable with laser welds, yet at the same time having an improved tolerance to joint fit-up when compared with laser welding.



Figure 9.11 A laser-hybrid welding apparatus.

When HLAW is used for welding metallic objects, the laser beam is focused to obtain intensities of more than 1 MW/cm^2 . When the laser beam hits the surface of the material, this spot is heated up to vaporization temperature, and a vapor cavity is formed in the weld metal due to the escaping metal vapor. This is known as a **keyhole**. The extraordinary feature of the weld seam is its high depth-to-width ratio. The energy-flow density of the freely burning arc is slightly larger than 100 kW/cm^2 . Unlike a dual process where two separate weld processes act in succession, HLAW may be viewed as a combination of both weld processes acting simultaneously in one and the same process zone. Depending on the kind of arc or laser process used, and depending on the process parameters, the two systems will influence each other in different ways. The combination of the laser and arc process results in an increase in both weld penetration depth and welding speed (as compared to each process alone). The metal vapor escaping from the vapor cavity acts upon the arc plasma. Absorption of the laser radiation in the processing plasma remains negligible. Depending on the ratio of the two power inputs, the character of the overall process may be mainly determined either by the laser or by the arc. Absorption of the laser

radiation is substantially influenced by the temperature of the workpiece surface. . In the hybrid process, the arc heats the metal, helping the laser beam to couple in. After the vaporization temperature has been reached, the vapor cavity is formed, and nearly all radiation energy can be put into the workpiece. The energy required for this is thus determined by the temperature dependent absorption and by the amount of energy lost by conduction into the rest of the workpiece.

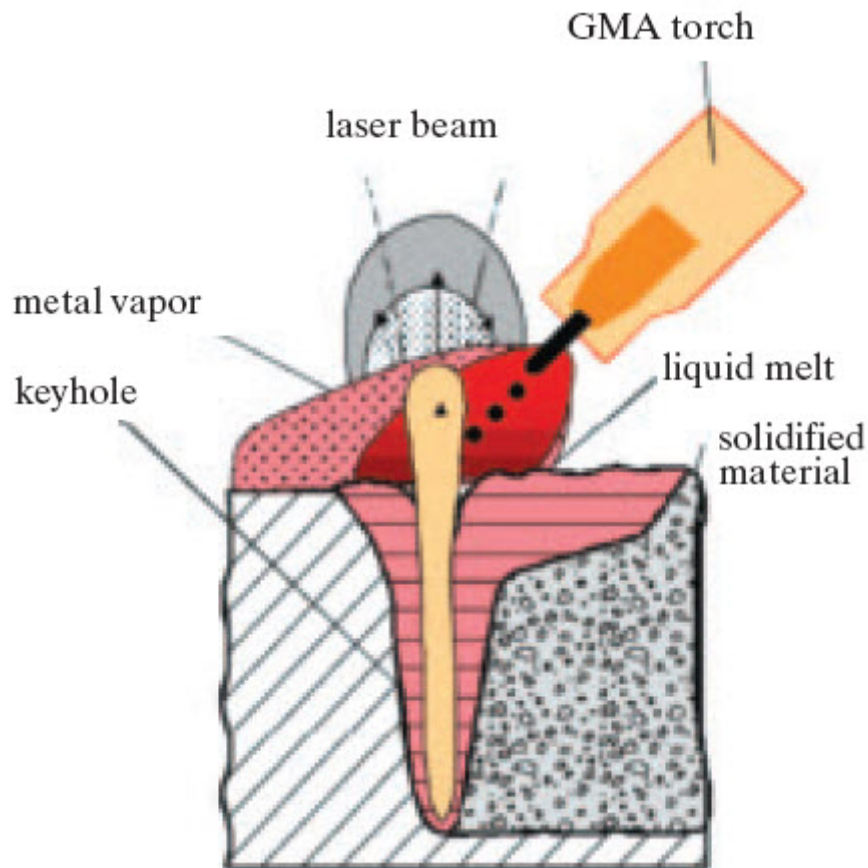


Figure 9.12 A schematic representation of a laser-hybrid welding.

HLAW can be used for a wide range of materials, including C-Mn steels, stainless steels and aluminium alloys, and less commonly titanium and nickel alloys. Typical joint geometries include butt and fillet welds. The following figure shows the differences between MiG weld, an Nd:YAG laser weld, and a HLAW weld.

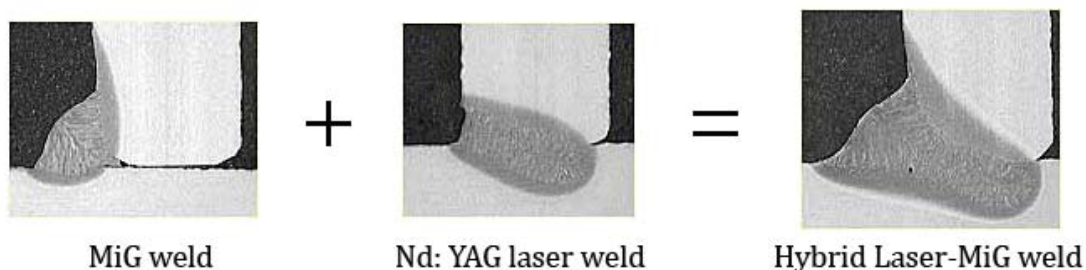


Figure 9.13 A metallographic cross-section of a laser-hybrid weld.

The technology was originally developed to provide a more cost-effective method for producing beams from high-strength plate for use on U.S. Navy ships; so

as it turns out the HLAW process started out as a shipbuilding-oriented welding method. The HLAW process can weld beams from ordinary and high-strength plate at rates that are five to ten times that of conventional welding processes and has demonstrated that these laser-welded structures perform as well as, or better than, conventionally welded structures in all respects. Additionally, tests have shown that the laser welds can achieve fatigue lives that are two to three times as long as conventional GMAW and SAW processes. HLAW is also used in fabricating steel sandwich panels. Sandwich panels can be simply defined as a three-layer structure consisting of two thin outer skins of high-strength material, separated by a low-density and low-weight core material. This creates a panel with excellent stiffness and weight characteristics. In the past, the development of all-metal sandwich structures has been inhibited by the limitations of conventional welding processes; HLAW techniques have finally made such structures practical.

In ship applications, the combination of laser/GMAW capabilities has enabled the reduction of beam structure weight 20 to 50% and is projected to substantially reduce structural fabrication and assembly costs. It has been projected that panel line velocity, a limiting factor in shipyard throughput, could be increased 38–49% by taking advantage of this technology for plate butt joint welding and stiffener-to-deck welding.

One major limitation of the HLAW process, however, is that before the laser welding process can start, the initial reflectance of the material must be overcome, especially on aluminum surfaces. This can be achieved by preheating the material.

DISSIMILAR WELDING OF STAINLESS STEELS

10.1 DISSIMILAR METAL WELDING ^[2] ^[51] ^[52]

With raw-material costs on the rise, selecting metals for a particular product has become increasingly difficult. What used to be a stainless design throughout now may be made up of several different alloys, with the most expensive ones used only where needed. With some alloys quadrupling or even quintupling in price in recent years, such design changes make absolute sense.

As a result, ***Dissimilar Metal Welding (DMW)*** is frequently used to join stainless steels to other materials. This approach is often used when a transition in mechanical properties or performance in service are required. However, in order to be welded together the metals need to have mutual solubility, and the intermetallic compounds they create must resist cracking, corrosion and other causes of weld failure. For example, power plant piping used to consist mainly of stainless steel and most probably of 309L steel for high-temperature applications. This is not the case anymore, since the rising cost of stainless steel has forced engineers to use stainless where susceptibility to corrosion or pressure requires it. Yet as soon as it isn't required, designers shift to other less costly materials, even finally transitioning down to mild steel.

Cladding is another form of DMW, putting strength and corrosion resistance where it's needed most, while saving costs elsewhere. It has become increasingly common to see a stainless cladding over ***high-strength low-alloy (HSLA)*** steel. This gives high corrosion resistance, thanks to the stainless steel, and high strength, thanks to the HSLA steel, allowing engineers to use thinner steel members. The joint between the clad and base metal is a dissimilar weld itself, and welders joining two clad materials together must consider the effects on all materials involved. In most cases these joints are welded from both sides, with weld passes on the substrate side and clad side, and the clad material removed through carbon-arc gouging. This

is evident on the following figure, showing methods of joining clad to base metal. From a dissimilar weld perspective, it's the clad side that presents the most challenges when choosing a filler metal.

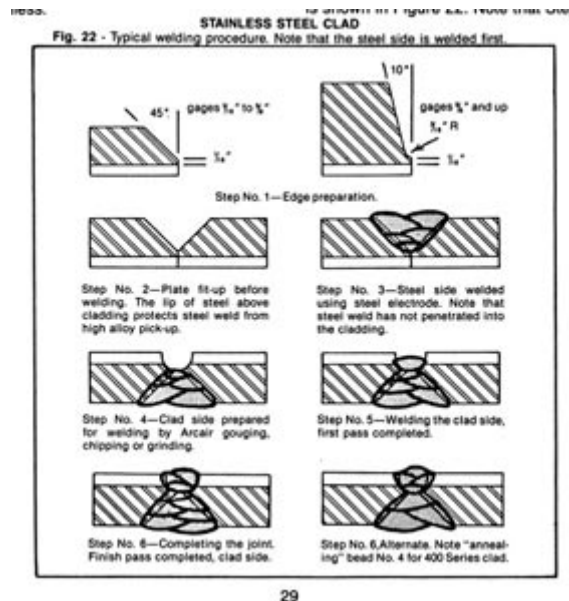


Figure 10.1 Examples of troublesome welds in cladding.

It should be noted, however, that the same methodology applies here as with other dissimilar metal welds. The resulting weld must have equal or greater corrosion resistance than the clad material and any filler material used must provide a suitable transition or buffer between differing melting points, thermal expansion and solubility characteristics.

In general, one stainless steel may be joined to another stainless steel, with varying degrees of dissimilarity. Since stainless steels can have martensitic, ferritic, austenitic and duplex microstructure, there are a vast number of microstructural combinations possible and the reasons for such combinations may be economic, property considerations, or sometimes just convenience. Stainless steels may also be welded to nickel-base alloys in demanding applications, where a transition in properties is required to accommodate more corrosive environments or there are superior strength requirements at elevated temperature. Finally, stainless steels are also sometimes joined to other non-ferrous alloy systems, such as aluminum and copper and these welds are usually made with a solid-state welding process.

10.2 SCHAEFFLER AND WRC-1992 DIAGRAMS CONSTITUTION PREDICTION^{[2] [51]}

When welding stainless steel to dissimilar metals, in general, control of the weld metal microstructure, particularly in the root-pass, is of critical importance since weld deposits can range from fully martensitic to fully austenitic, or exhibit in-between mixtures. This is why differences in physical and mechanical properties between the weld and base metals will almost certainly exist. For example, differences in the coefficient of thermal expansion may result in locally high stresses that can promote service failures, particularly due to thermal cycling from low to high temperatures.

That, however, is not the sole reason why control of the weld metal's constitution during the root-pass (first weld pass) is so important; dilution effects are greatest in this pass. Low-alloy and carbon steels contain rather high carbon content, and carbon tends to be an issue in any welding scenario. As it has been already described, chromium reacts with carbon forming chromium carbide, a material with very high hardness and low ductility. These are both extremely deleterious for any weld. Its melting temperature is also very high, so it solidifies before the rest of the weld pool. As the remaining weld solidifies, the chromium carbides migrate to the middle of the weld, at the grain boundary. Solidified at the boundary, the metal becomes extremely brittle and can sometimes break, since the material does not have the ductility required to withstand the solidification stresses around it.

A solution for all the aforementioned problems is selecting a proper filler metal. If the filler metal contains enough nickel, it will dilute the carbon, forming a suitable transition between the low-alloy and stainless steel. In general, should the filler metal give stable austenite with a small amount of ferrite in the first weld pass, weld solidification cracking is very unlikely. Nickel can also solve many problems with higher-ferrite materials held at high temperatures for long periods. Instead of diluting the carbon and stainless steel with a ferritic base, the welder simply uses a nickel base, assuring that weld deposit will be austenitic and endure prolonged exposure to high temperature. As shown in the following figure, as Ni content within the filler metal increases, the overall Ni content of the weld metal increases substantially.

	304(L) Stainless	Carbon Steel	309(L) Filler	Weld Deposit
C	0.02	0.2	0.03	0.06
Mn	1.33	0.9	1.5	1.35
Si	0.67	0.35	0.6	0.56
Cr	19	0	23.5	17.9
Ni	10	0	13	9.8

	304(L) Stainless	Carbon Steel	Ni Filler (NiCrFe)	Weld Deposit
C	0.02	0.2	0.04	0.06
Mn	1.33	0.9	3	2.2
Si	0.67	0.35	0.3	0.3
Cr	19	0	20	15.8
Ni	10	0	72	45.2

The Schaeffler diagram and the WRC-1992 diagram, which is modified to include boundaries for martensite transformation, can both be used to predict weld metal composition, as well as the composition of a narrow transition region that exists between the bulk weld metal and the base metal and has distinctly different microstructure and properties from the adjacent regions. An example as to how these diagrams can be used is shown in the following figure.

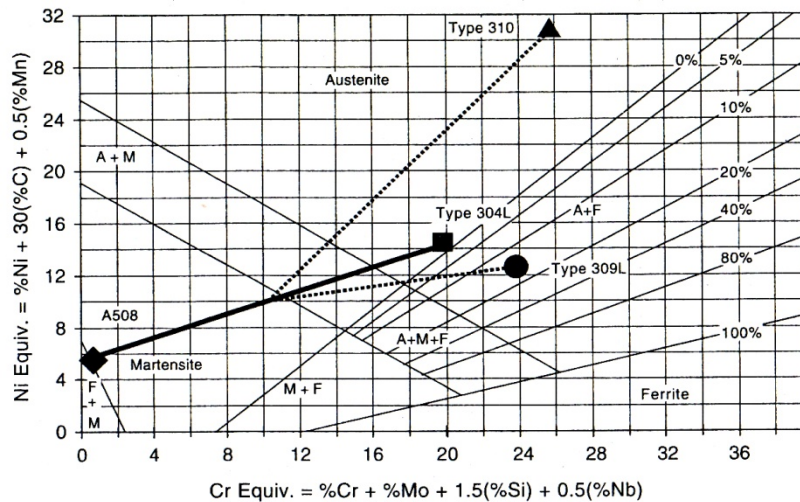


Figure 10.2 Prediction of weld metal constitution using the Schaeffler diagram.

A low-alloy steel such as ASTM A508 is being welded to Type 304L, using either a Type 309L or Type 310 filler metal. The intersection of the dotted lines along the tie-line between the two base metals represents equal dilution of the base metals by the weld metal composition. The composition represented by the intersection can be viewed as a synthetic base metal composition. Assuming there is equal base metal dilution, the ultimate composition of the first pass weld metal will fall along the dotted tie-line connecting this point to the filler metal composition. If the total base metal dilution into the first pass or filler metal is 30%, which is normal for welding with covered electrodes, the weld metal composition will lie along this second tie-line, 30% of the distance from the 309L point toward the synthetic base metal composition. That composition is approximately the place where this second tie-line crosses the isoferrite line labeled “5% ferrite”. Subsequent passes will lie closer to the filler metal composition. Note that for the 309L filler metal, normal dilution of the filler metal by the base metal will actually result in a two-phase austenite + ferrite microstructure, which is usually highly resistant to weld solidification cracking.

The WRC-1992 diagram, modified to include boundaries for martensite transformation at 1% Mn, is used in the exact same way.

However, it should be noted here that predicting the transition region microstructure can be extremely difficult, since this region may change dramatically over a very short distance (about 1 mm or 0.04 in.). A perfect example of such compositional transition is **carbon diffusion (migration)** from the HAZ to the fusion zone during welding or PWHT. This usually happens when the base metal has higher carbon content than the weld metal (which is usually the case) and potentially results in narrow martensitic region at the fusion boundary that exhibits high hardness. Should the weld metal be high in chromium and the base metal have little or no chromium, the tendency for carbon migration from the HAZ into the weld metal during PWHT is greater. Carbon migration has been observed from carbon steel base metals across the fusion boundary and into the aforementioned transition region and can result in local microstructural changes in both the HAZ and the transition region. The result is a carbon-depleted zone in the HAZ, immediately adjacent to the fusion boundary. Soft ferrite forms in this carbon-depleted zone, which has been known to lead to premature failure in creep.

In situations where the base metal is ferritic at temperatures near the melting point and the weld metal is austenitic, normal epitaxial growth may be suppressed. This can result in the formation of **Type II boundaries**, which run roughly parallel to the fusion boundary, as seen in the following figure. Type II boundaries are in contrast to the **Type I boundaries**, which result from columnar growth from base metal grains into the weld metal and run roughly perpendicular to the fusion boundary.

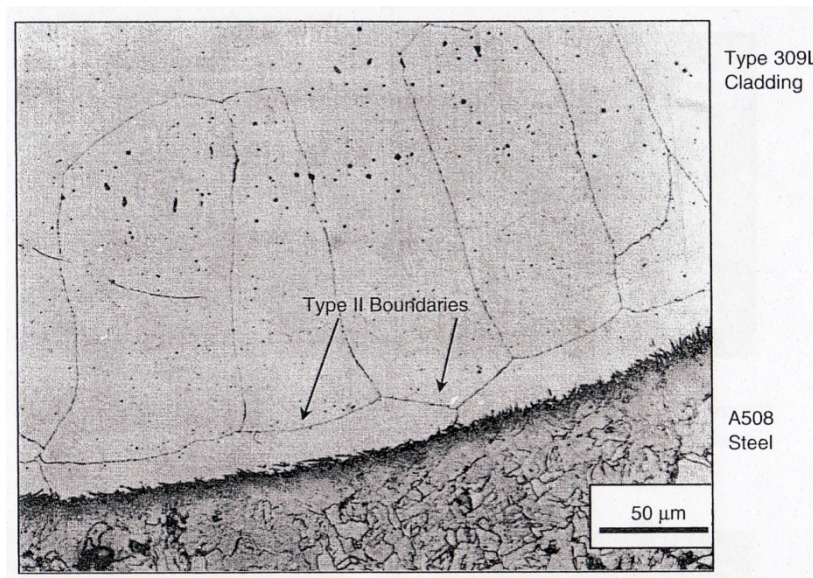


Figure 10.3 Orientation of Type II boundaries.

Type II boundaries are clearly evident just adjacent to the fusion boundary in the weld metal. They form in the solid state during weld cooling when both the weld metal and HAZ are austenitic, allowing austenite grain growth across the fusion boundary. These boundaries may also be sites for carbide precipitation. Type II boundaries are of special interest because they have been associated with a number of instances of in-service cracking, also called **disbonding**. Disbonding can occur along these boundaries, particularly in situations where stainless steel cladding is applied to carbon steels. Due to the mechanism of Type II boundary formation, the substrate of the clad metals transforms to ferrite or other decomposition products, and the compositional transition zone, or part of it, transforms to martensite. This martensite may extend to the Type II boundary or stop short of it, as shown on the following picture.

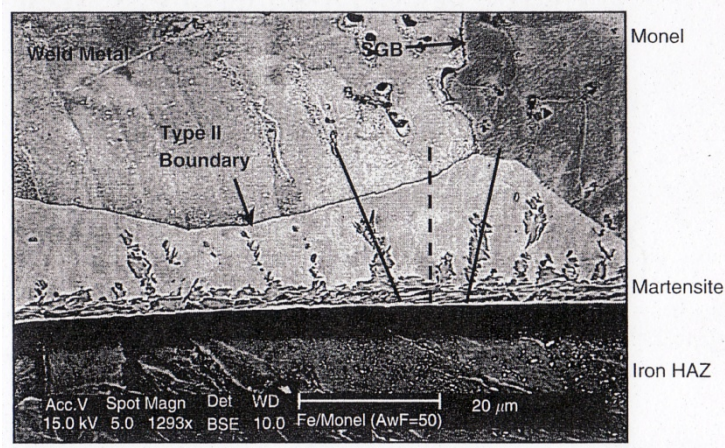


Figure 10.4 Creation of a martensite substrate in a case of clad disbonding.

Because of these interchangeable formations there is a mismatch in the coefficient of thermal expansion between the BCC or martensite side of the transition region and the FCC cladding side. Thermal cycling sets up strain in this region and any Type II boundaries are weak planar interfaces, which makes them preferential cracking sites. In addition, the martensite is a potential location for hydrogen-induced cracking if the service environment produces hydrogen charging.

10.3 WELDABILITY AND ISSUES^{[2] [52] [53]}

Dissimilar metal welds between carbon steels and stainless steels are associated with a number of cracking mechanisms. These include solidification cracking, clad disbonding along Type II boundaries, creep failure in the HAZ of the carbon steel, hydrogen-induced cracking along the fusion boundary and reheat or PWHT cracking.

Solidification cracking most commonly occurs when nominally austenitic filler metals, such as Type 308L and 309L, are used. If the root pass weld deposits are heavily diluted by the carbon steel, they will solidify as primary austenite because there was insufficient ferrite potential in the filler metal and/or there was excessive dilution. The following figure shows an example of solidification cracking in a 309L submerged arc fillet weld between A36 steel and 304L stainless steel.

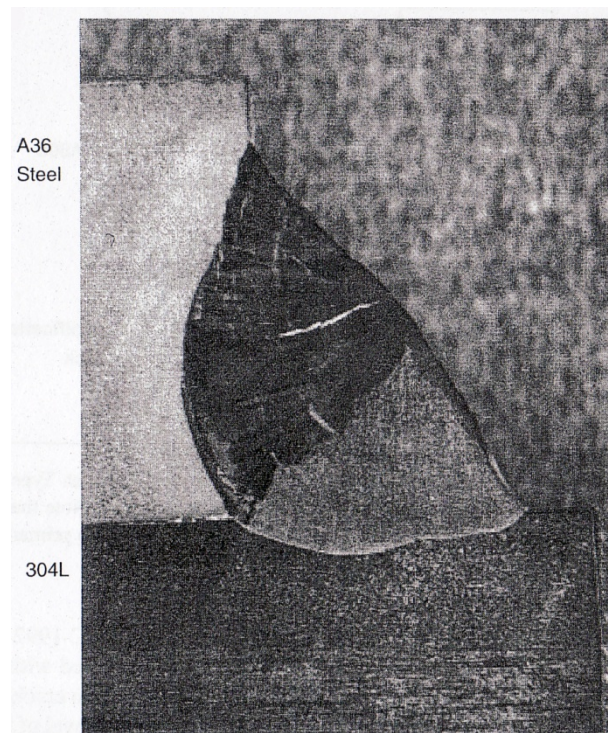


Figure 10.5 Solidification cracking in a 309L submerged arc fillet weld between A36 steel and 304L stainless steel.

The *light etching* appearance reflects metal that solidified as *primary ferrite*, while the *darker etching* appearance reflects metal that solidified as *primary austenite*. It should be noted that the solidification crack lies along the centerline, not of the entire weld, but of the region of primary austenite solidification. Better control of the weld process (torch positioning and heat input) can be very successful in ensuring FA solidification and minimizing any risk of cracking. In short, solidification as primary austenite (AF) or in the fully austenitic mode (A) is bound to cause centerline solidification cracking.

Clad disbonding typically occurs along Type II boundaries in the clad deposit. Although the precise mechanism for this type of failure is still not known, it is clear that the nature of this boundary and its presence in a microstructural and compositional transition zone leads to these types of failures. Weld deposits that do not contain Type II boundaries are immune to this type of failure. Clad disbonding may occur during the actual cladding operation, during subsequent PWHT, or in service. It is often difficult to know when the actual failure occurs, since inspection is

usually not carried out until after the PWHT is completed. Unfortunately, there is no way to avoid Type II boundary formation in most common dissimilar combinations.

Hydrogen-induced cracking along the fusion boundary has many times been considered the prime “suspect” for clad disbonding. There have been numerous reports for such failures, most of which have been observed in cladding applications where a corrosion resistant austenitic alloy is applied to a ferritic structural steel. As already discussed, however, this form of cracking has occurred during fabrication, prior to exposure to a hydrogen environment. The fact that disbonding can occur without prolonged exposure to hydrogen suggests, among other things, that hydrogen absorbed during welding can cause cracking near the dissimilar metal fusion boundary. As for the mechanics for this form of cracking, it has been generally reported that hydrogen-induced cracking typically follows Type II grain boundaries and as a result cracking is nominally located within the martensitic transition zone near the fusion boundary, where Type II boundaries often reside. It should be noted here that numerous experiments have come to the conclusion that cracking is always confined to regions containing martensite and there severity in filler metals is a function of their minimum dilution to form martensite on the Schaeffler diagram; filler metals with the lowest minimum dilution to form martensite also presented the most severe cracking. Another point of interest is that cracking was never observed in welds made with pure argon shielding gas.

Creep failure in the HAZ of the carbon or low-alloy steel is a type of failure that has been observed in thick-section structural welds, just adjacent to the fusion boundary. Carbon migration from the HAZ to the weld metal during welding, PWHT, or service exposure results in a soft ferritic HAZ microstructure. Under imposed residual and thermal stresses, creep cracks can occur along the ferrite grain boundaries. As a result, this region becomes subject to creep failure, under elevated temperature exposure. However, it is most common for these failures to occur after 10 to 15 years of elevated-temperature service. One other thing that should be reported is the significant difference of the average linear CTE between austenitic stainless steel and carbon or low-alloy steel, with the former being greater than the latter. This difference in CTE can result in large strains when a dissimilar weld is heated to an elevated temperature. Because of the carbon steel’s lower CTE, it will try to constrain the stainless steel weld from expanding. This results in locally high stresses at the interface. Over time, grain boundary sliding occurs, eventually leading to a creep failure in the carbon steel HAZ. The use of Ni-base filler metals has been, as always, very effective in reducing the thermally induced CTE mismatch stresses, thereby avoiding cracking.

Reheat or PWHT cracking is generally defined as intergranular cracking in the HAZ or in the weld metal that occurs during the exposure of welded assemblies to the elevated temperatures produced by PWHT or high temperature service. While trying to relieve these residual stresses through PWHT the relaxation strains may actually exceed the material’s local ductility, thus resulting to local cracking. Factors influencing either the level of residual stress resulting from welding, or the deformation mechanism involved in relieving them, are therefore significant. Reheat cracks were reported to occur in the HAZ of low-alloy constructional steels, including the quenched and tempered grades used for pressure vessel applications. They appeared as small cracks in the HAZ under the austenitic cladding, as witnessed in the following figure, or as microcracks in the HAZ of the main structural welds.

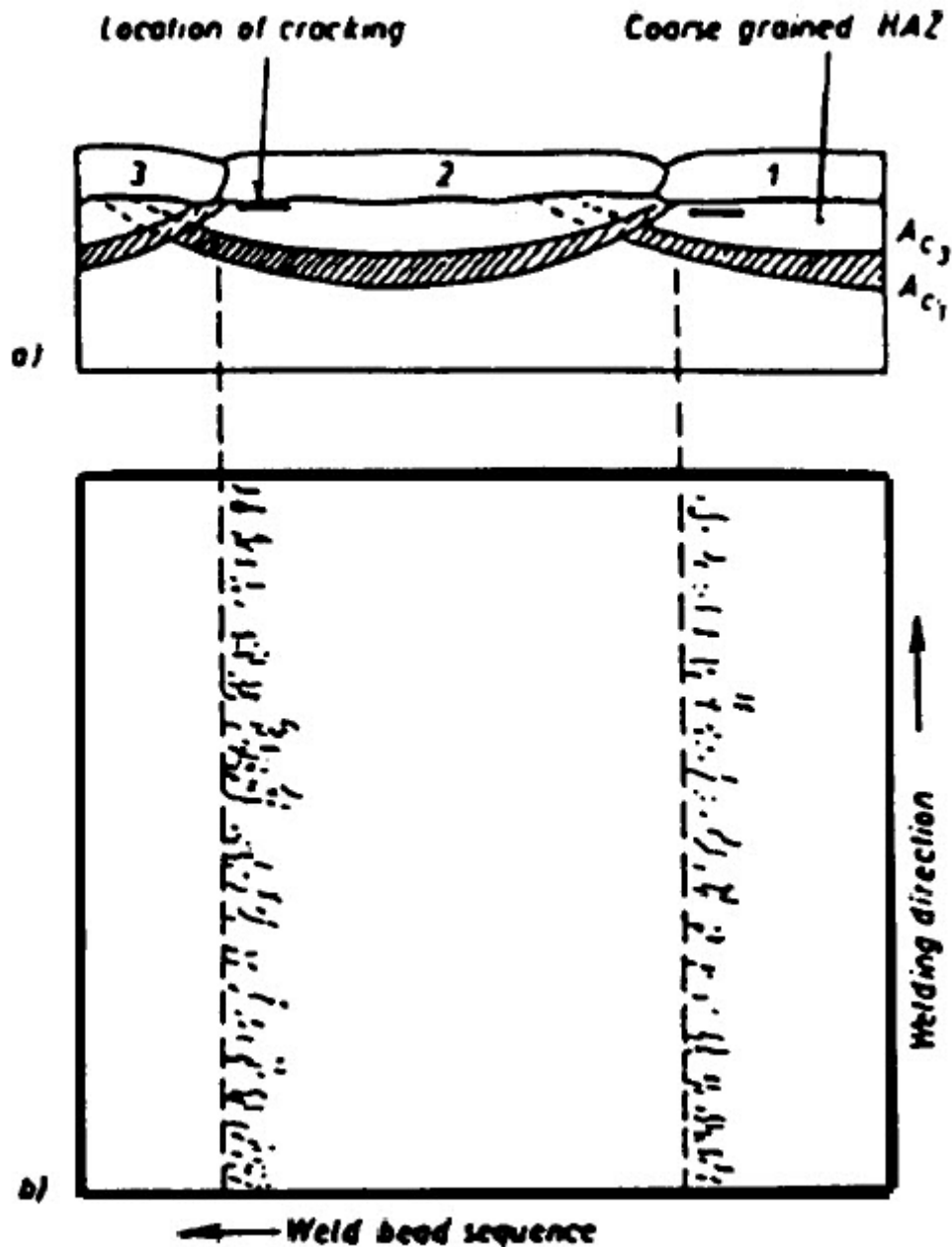


Figure 10.6 Schematic representation of a reheat cracking defect.

This type of cracking has necessitated a high level of inspection and extensive experimentation. Early investigations on the mechanism of reheat cracking studied the influence of alloying elements, which lead to secondary precipitation hardening and came to the conclusion that their presence is indeed deleterious. However, the embrittlement effect of each impurity element and the amounts in which they can be tolerated are in contention. Although not conclusive, these studies have at least highlighted possible ways of obtaining improvement in reheat cracking behaviour in certain low-alloy steels by realistic control of composition, e.g. by stabilizing sulphur with titanium, zirconium and particular rare earth metals.

Dissimilar metal weld failures are also some times attributed to the very sharp changes in composition and corresponding properties, which occur along the fusion line of the weld. Sharp nickel and chromium composition gradients develop near the fusion line of the DMW due to partial mixing between the two materials that occurs

during welding. These compositions gradient, combined with the high cooling rate during welding, form a hard and brittle martensitic region in the partially mixed zone of the weld. Stainless steel alloys typically have lower carbon levels than the alloy steels. This leads to a large carbon concentration gradient across the DMW joint, and this, in turn, results in a significant localized reduction in the creep strength on the carbon steel side of the weld and a modification of the microstructure during heat treatment. The sharp changes in hardness and strength described above occur over very short distances and are combined with significant differences in thermal expansion between the two materials. Although the use of Ni-base filler metals and joint designs with wide included angles can extend the DMW's life, as discussed earlier, they do not provide a long-term solution to the problem, because failures still occur in joints prepared with these modifications. This can be rectified by a process referred to as ***Laser Engineering Net Shaping (LENS)***, through which a transition joint between carbon steels and stainless steels can be prepared. This process appears to be well suited for fabricating carbon steel-to-stainless steel transitions joints in which composition is varied in a controlled manner over relatively large distances.

10.4 VARIOUS DISSIMILAR COMBINATIONS

10.4.1. Welding stainless steels to various metals. ^[56]

Most stainless steels can be successfully welded to mild- and low-alloy steels. Considerations must be given to effects of dilution of the weld metal with the two base metals, as well as to the different coefficients of thermal expansion of stainless steel and mild- or low-alloy steels. The weld metal deposited will tend to pick up alloys from both parts of the joint. The effect of this dilution can be controlled by buttering or overlaying the surface of one of the metals being joined and by selecting the correct electrode or filler material. The weld joint between a slightly ferritic stainless steel and a mild- or low-alloy steel would be hard and brittle if made with a slightly ferritic stainless steel electrode. However, if a fully austenitic stainless steel electrode or filler rod is used, the amount of ferrite in the weld metal would be reduced to a tolerable level.

As for the thermal expansion coefficients, austenitic stainless steel has a coefficient of thermal expansion about twice that of any mild- or low-alloy steel. During the cooling cycle of the weld, the stainless steel side will tend to contract more than the mild- or low-alloy steel side. This difference in contraction will set up stresses in the weld joint. If the weld joint is subjected to repetitive thermal cycles, the resulting stress cycling could cause premature failure similar to fatigue fracture. The use of stainless steel for buttering will not solve this situation. The buttering technique is not recommended for those situations in which repetitive thermal cycles are involved.

When welding steel to different types of steel, however, things are somewhat different. These steels are normally welded with electrodes matching the composition of the lower-strength material. This applies not only to various grades or strength levels of carbon and low-alloy steels, but also to various grades of stainless steels.

Various examples of stainless steel dissimilar welding will be discussed and explained in detail below and recommended solutions will be given in most cases.

10.4.2. **Nominally austenitic alloys whose melted zone is expected to include some ferrite or to solidify as primary ferrite** ^[2]

This joint has perhaps the least dissimilarity and the least potential difficulty. Examples include joining 304L to 316L or 304H to 347. There is generally wide latitude in selecting filler metal. The least expensive or most readily available filler metal that matches one or the other base metal is likely to be quite suitable. From the point of view of serviceability of the joint, it matters little whether the filler metal is 308L or 316L for the 304L to 316L joint. Similarly, it matters little whether the filler metal is 308H or 347 for the 304H to 347 joint. Furthermore, an overalloyed filler metal, such as 309L, would also be suitable for the 304L to 316L joint. There should be no weld solidification cracking issues in such joints.

10.4.3. **Nominally austenitic alloys whose melted zone is expected to contain some ferrite, welded to fully austenitic stainless steel.** ^[2]

Examples include 316L welded to a superaustenitic 6% Mo stainless steel, or 304H welded to Type 310 stainless steel. In these situations there can be a risk of welded solidification cracking if the weld metal composition is such that it solidifies as primary austenite, but a weld deposit with appreciable ferrite may be

unacceptable due to corrosion considerations in the first example or sigma phase formation concerns at high temperature in the second example. If ferrite in the weld is acceptable in the 316L to 6% Mo steel joint, an intermediate-alloy filler metal such as 309LMo is usually the best choice. But if ferrite is not acceptable, a low-ferrite 316L filler metal or Type 385 filler metal, both of which have good solidification cracking resistance despite the lack of ferrite, is likely to be the best choice. In the 304H to 310 joint, a 308H or 309 filler metal may be acceptable if high-temperature service is not anticipated, but this combination of base metals is likely to be used at high temperature. In this case, a 310 filler metal may be a better choice than a 308H filler metal, because without ferrite the weld metal has reasonably good solidification cracking resistance if applied correctly. In cases where the weld metal is expected to be fully austenitic, correct application includes: (1) small, relatively low-heat-input weld beads; (2) a convex profile; (3) care to fill craters.

10.4.4. Austenitic stainless steel joined to duplex stainless steel. ^[2]

An example of this situation is joining 316L stainless steel to 2205 duplex stainless steel. In this case, a filler metal selection to obtain ferrite is easy. Either 316L filler metal or 2209 filler metal is acceptable, and the choice is made based on cost and availability. On the other hand, 309L filler metal would also be expected to provide ferrite, but because 309L contains no molybdenum, it might be an inappropriate choice. The two base metals contain Mo, which makes them more resistant to localized corrosion (particularly pitting) in some environments relative to Mo-free 309L.

10.4.5. Austenitic stainless steel joined to ferritic stainless steel. ^[2]

An example of this situation is joining Type 444 stainless steel to Type 316L stainless steel. Because ferritic stainless steel filler metals are not readily available, the most common choice, with acceptable results, is likely to be the filler metal that is appropriate for the austenitic base metal, 316L in this case.

10.4.6. Austenitic stainless steel joined to martensitic stainless steel. ^[2]

The filler metal of choice depends on whether or not the weldment will subsequently see PWHT or elevated-temperature service. If no elevated temperature exposure is expected after welding, a filler metal expected to produce weld metal with primary ferrite solidification is usually acceptable. For example, 410 stainless steel might be joined to 304 stainless with 309 filler metal. However, should elevated-temperature exposure be expected, a crack-resistant fully austenitic filler metal such as 310 would be a more appropriate selection.

10.4.7. Martensitic stainless steel joined to ferritic stainless steel. ^[2]

An example might be welding Type 410 to Type 409 stainless steel. One might be tempted to select either a ferritic stainless steel filler metal or a martensitic stainless steel filler metal. That may be acceptable if mechanical property demands on the joint are not high. Dilution of the filler metal with both base metals is likely to produce a weld metal of mixed ferrite-martensite microstructure. Ordinarily, a mixed ferrite-martensite weld metal microstructure has poor ductility, and if a ductile weld is required, an austenitic stainless steel filler metal like 309L, or a nickel-base alloy filler metal like NiCr-3 may be more appropriate. If high-temperature thermal cycling is likely, the nickel-base alloy filler metal may be a better choice than the 309L, because of a better match in the coefficients of thermal expansion.

10.4.8. **Stainless Steels joined to “Difficult-to-Weld” Steels.** [2] [48] [50]

Stainless steel filler metals are often used for joining “Difficult-to-Weld Steels” to stainless steels. “Difficult-to-Weld Steels” are alloy steels with more than 0.25wt% C. An example of such high carbon steel used in shipbuilding is the armor of naval vessels. **Krupp cemented armor**, the armor applied on today’s naval vessels, has a typical chemical analysis as follows:

Carbon:	0.35
Nickel:	3.90
Chrome:	2.00
Manganese:	0.35
Silicon:	0.07
Phosphorus:	0.025
Sulphur:	0.02

Carbon being the principal hardening element, it was only natural to carry its content as high as possible, even if something like that impeded its manufacture, forging and fibering and caused the plates to become brittle. These drawbacks were rectified by the proper addition of nickel and chrome.

Such high carbon content, however, (and any carbon content above 0.04wt%, for that matter) warrants special concern regarding the formation of chromium carbides at the grain interface. This transformation does not really require any preheat treatment. Austenitic stainless steels, in particular, are prone to hot cracking and this is why welding carbon to stainless steels requires that the stainless weld metal has sufficient ferrite to resist cracking.

In welding carbon steels to stainless steels, the austenitic and nickel filler metals offer unique features that can reduce weld crack potential in both the welds and HAZ’s. High carbon, high strength steels welded to each other are subject to hydrogen-induced cracking. The three fundamental requirements for hydrogen-induced cracking are:

1. High hardness.
2. A source of hydrogen.
3. High stresses.

With the high carbon steels, high hardness is more or less typical in the HAZ, unless very high preheat and interpass temperatures are utilized for the welds. Hydrogen, on the other hand, can be derived from many sources and most times is practically impossible to avoid. Finally, the stresses that can influence HAZ cracking typically result from weld residual stresses caused by weld shrinkage. These stresses can be further exaggerated by weld joint restrictions, as found in certain fixtures.

The austenitic or nickel filler metals greatly reduce the weld transverse cracking potential and they also reduce, slow down or trap the hydrogen that can diffuse from the weld to the HAZ. As a result, the potential for hydrogen-induced cracking in the HAZ is significantly reduced. The diffusion of hydrogen through austenitic and nickel filler metal steels and welds can be almost 80-110 times slower than through carbon steels and welds, however they can still absorb hydrogen so these electrodes should be treated with the same respect and rules that apply to low hydrogen filler metals.

When welding carbon to stainless steels using the MIG welding method, a gas solution of 98% Ar - 2% CO₂ (*Ed's unique Stainless, Duplex MIG Gas Mix*) or 90% He – 7.5% Ar – 2.5% CO₂ should be used to further reduce any type of weld cracking.

The Schaeffler diagram and the WRC-1992 diagram, modified to include martensite boundaries, can be used to predict whether or not the root-pass weld metal will contain enough ferrite to resist cracking, and whether or not the root-pass weld metal will transform to martensite. The two diagrams, however, do not always make similar predictions and in situations like this it is the WRC-1992 diagram that has been shown to make the more accurate predictions.

When welding high carbon steels to stainless steels a 309L filler metal is mainly used and the resulting ferrite is approximately 14-16 FN. When using a 309L wire for a weldment like this, however, the first weld pass is likely to end up with “insufficient ferrite”. The reason for this is that the carbon from the high-carbon steel, when mixed with the stainless steel weld, will suppress the ferrite formation. So instead of the 309L for this application, a 312 electrode may be recommended. This filler metal produces much higher ferrite levels than the 309L and that is why the 312 filler metal is recommended for applications sensitive to weld cracks. Filler metals such as 307, 308Mo and 310 can resist cracking with the aid of alloys and without the aid of ferrite.

10.4.9. Copper-base alloys joined to stainless steels ^[2] [63]

When welding copper, in general, special notice should be given to its high thermal conductivity, which contributes to dissipate rapidly most of the heat that should be used up for local melting, although this property varies widely among the different alloys. Copper alloys are usually selected for their high electrical and thermal conductivity, for their pleasant appearance and color with plating and finishing, for their ease of fabrication, forming and machining, for their corrosion resistance, and absence of magnetism.

However, extreme caution should always be exercised, since most copper alloys include volatile and toxic elements that are liberated to the atmosphere during welding. As a consequence, it is imperative to enforce special precautions in order to collect and dispose of noxious fumes, powders and dust and to provide to the welder an effective ventilation system.

Another offset is copper's tendency to wet austenite grain boundaries in stainless steels, which can lead to severe cracking problems if stainless steel is deposited on copper, or if copper is deposited on stainless steel. As a result, a non-matching filler metal is usually chosen for such joints. Probably the most common is the over 90% Ni filler metals of AWS classes ENi-1 (covered electrode) or ERNi-1 (bare wire or rod). Small convex weld beads, with attention to filling craters, are usually recommended.

Special consideration should also be given to the welding process that one uses for joining copper-base alloys to stainless steel; experience has shown that the most appropriate processes are **friction welding** and **electron beam welding**. Friction welding, in particular, is famous and almost unique in permitting welds of copper to aluminum or to stainless steel.

10.4.10. Nickel-base alloys joined to stainless steels. ^[2]

Because there are so many stainless steels and so many nickel-base alloys, possible combinations from the two alloy systems are very numerous. Problems with

solidification cracking often result when stainless steel filler metals are attempted for these joints. As a result, most joining of stainless steel to nickel-base alloys is done with nickel-base alloy filler metals. In most cases, the nickel alloy filler metal appropriate to joining the nickel-base alloy to itself is also appropriate for joining the nickel-base alloy to the stainless steel. For example, if the Ni-Cr-Mo-W alloy UNS N 10276 (commonly known as Alloy C-276) were to be joined to 316L stainless steel, a suitable filler metal would be ENiCrMo-4 or ERNiCrMo-4, which is a composition match for the C-276 alloy. Certain nickel-base alloy filler metals have a well-earned reputation for being able to join a wide variety of nickel-base alloys to a wide variety of stainless steels with great success. A perfect example is the NiCrMo-3 filler metal.

However, it is not possible to address each and every possible combination of stainless steel to nickel-base alloy joining. As a general guideline, one should always refer to the nickel-base alloy filler metal supplier's recommendation for combinations the filler metal could be used in.

10.5 STAINLESS STEEL FILLER METALS^[2] [48]

In general, stainless steel filler metals are widely used in a full spectrum of applications. The type of filler metal to be used is not only dependent on the base and weld metal, but on the properties required from the weldment as well as the service conditions expected for the weldment. A general guideline for stainless steel filler metal selection is the following:

- **308L and 308LSi filler metals:** they are predominantly used on austenitic stainless steels, such as types 301, 302, 304, 305 and cast alloys CF-8 and CF-3.
- **308L and 308H filler metals:** they are used for high temperature applications, but 308H electrodes will provide superior creep resistance than the one offered by 308L electrodes.
- **309L and 309LSi filler metals:** use them when joining 309 or mild steels and low alloy steels to stainless steels. They can also be used for some 308L applications. Since it does not contain molybdenum, however, it cannot be used for 316L and 316 applications.
- **309 filler metals:** use them when joining dissimilar stainless steels, such as 409 to itself or to 304L stainless steel.
- **316L and 316LSi filler metals:** these should be used with 316L and 316 base metals.
- **347 filler metal:** it is used for 347 and 321 base materials, because it matches these stabilized grades. It is also suitable for 308L filler metal applications.

When selecting from carbon and low-alloy steel welding consumables for SMAW of various metal combinations one could use the following table as a general guideline:

Base Material	Carbon Steel	Carbon-Molybdenum Steel	1 and 1 1/4 Cr - 1/2 Mo Steel	2 1/4 Cr-1Mo Steel	5Cr- 1/2 Mo Steel	9Cr-1Mo Steel
Carbon steel	AB	AC	AD	AE	AF	AG
Carbon-Molybdenum Steel		C	CD	CE	CF	CH
1 and 1 1/4 Cr - 1/2 Mo Steel			D	DE	DF	DH
2 1/4 Cr - 1 Mo Steel				E	EF	EH
5 Cr - 1/2 Mo Steel					F	FH
9 Cr- 1 Mo Steel						H

Legend:

- A AWS A 5.1 Classification E 70XX low hydrogen (E7018 preferred)
 B AWS A 5.1 Classification E6010
 C AWS A 5.5 Classification E70XX - A1, low hydrogen
 D AWS A 5.5 Classification E70XX-B2L or E80XX-B2, low hydrogen
 E AWS A 5.5 Classification E80XX-B3L or E90XX-B3, low hydrogen
 F AWS A 5.5 Classification E80XX-B6 or E80XX-B6L, low hydrogen
 G AWS A 5.5 Classification E80XX-B7 or E80XX-B7L, low hydrogen
 H AWS A 5.5 Classification E80XX-B8 or E80XX-B8L, low hydrogen

Notes:

- Table 1 refers to coated electrodes (SMAW process) only. For bare wire welding (SAW, GMAW, GTAW and FCAW), use equivalent electrode classifications (AWS A 5.14, A 5.17, A 5.18, A 5.20, A 5.23, A 5.28)
- Higher alloy electrode specified in the table should normally be used to meet the required tensile and toughness after post weld heat treatment (PWHT). If no PWHT is required, the lower alloy electrode specified may be required to meet the hardness requirements.

When selecting from austenitic, super-austenitic, and duplex stainless steel welding consumables, however, one should use the following table for advice:

Base Metal	304L SS	304H SS	316L SS	317L SS	904L SS	6% Mo SS	7% Mo SS	Alloy 20Cb-3	2304 Duplex SS	2205 Duplex SS
Carbon and Low Alloy Steel	ABC	ABC	ABC	ABC	ABC	ABC	ABC	ABC	N	N
Type 304L Stainless Steel	D	DE	DF	DG	DC	C	C	DCH	NL	NL
Type 304H Stainless Steel		E	EF	EG	*	*	*	ECH	*	*
Type 316L Stainless Steel			FG	FG	FC	FC	FC	FCH	NL	NL
Type 317L Stainless Steel				GC	GC	GC	GC	GC	L	L
Type 904L Stainless Steel					C	C	C	C	L	L
Type 6% Mo Stainless Steel Eg: 254 SMO, AL 6XN						CJK	CJK	*	*	*
Type 7% Mo Stainless Steel Eg: 654 SMO							CJK	*	*	*
Type Alloy 20Cb-3								H	*	*
Type 2304 Duplex SS									LM	LM
Type 2205 Duplex SS										LM

Legend

- A AWS A 5.4 Classification E309L-XX
 B AWS A 5.11 Classification ENiCrFe-2 or -3
 (-2 is Alloy 718 and -3 is Inconel 182)
 C AWS A 5.11 Classification ENiCrMo-3 (Inconel 625)
 D AWS A 5.4 Classification E308L-XX
 E AWS A 5.4 Classification E308H-XX
 F AWS A 5.4 Classification E316L-XX
 G AWS A 5.4 Classification E317L-XX
 H AWS A 5.4 Classification E320LR-XX
 J AWS A 5.11 Classification ENiCrMo-4
 (Hastelloy C-276)
 K AWS A 5.11 Classification ENiCrMo-11
 (Hastelloy G-30)
 L AWS A 5.4 Classification E2209-XX
 M AWS A 5.4 Classification E2553-XX
 N AWS A 5.4 Classification E309MoL-XX

* Unlikely combination

Notes:

- Table 2 refers to coated electrodes only. For bare wire welding (GMAW, GTAW), use equivalent electrode classification (AWS A 5.14)
- There are many proprietary alloys available in the market and material combinations you might encounter. Consult the manufacturer or the DSF for proper filler metal selection

DEFINITIONS

AL-6XN: the AL-6XN alloy is a superaustenitic stainless steel which was developed by Allegheny Ludlum Corporation. It exhibits far greater resistance to chloride pitting, crevice corrosion and stress-corrosion cracking than exhibited by the standard 300 series stainless steels, and is less costly than the traditional nickel-base corrosion resistant alloys.

Annealing: a heat treatment, wherein a material is altered causing changes in its properties such as strength and hardness. It is a process that produces conditions by heating to above the re-crystallization temperature and maintaining a suitable temperature, and then cooling. Annealing is used to induce ductility, soften material, relieve internal stresses, refine the structure by making it homogeneous, and improve cold working properties.

AOD: Argon Oxygen Decarburization

Argon-Oxygen Decarburization: is a process primarily used in stainless steel making and other high grade alloys with oxidizable elements such as chromium, aluminum, etc. After initial melting, the metal is transferred to an AOD vessel where it will be subjected to three steps of refining; decarburization, reduction, and desulphurization.

Butter weld: a weld caused of one or more string or weave beads laid down on an unbroken surface to obtain desired properties or dimensions.

Buttering: the act of producing a **butter weld**

Carburization: a heat treatment process in which iron or steel is heated in the presence of another material (but below the metal's melting point) which liberates carbon as it decomposes. The outer surface or case will have higher carbon content than the original material. When the iron or steel is cooled rapidly by quenching, the higher carbon content on the outer surface becomes hard, while the core remains soft and tough

Constitutional liquation: a condition in which melting of the interface between phases of multiphase alloys is possible, provided these alloys are rapidly heated.

CTE: coefficient of thermal expansion

DDC: ductility-dip cracking

Doping: doping is the process of intentionally introducing impurities into an extremely pure (also referred to as *intrinsic*) semiconductor to change its electrical properties.

Electrocleaning: an annealing process used to remove scale or other impurities from various alloys. An improved method of electrocleaning metal includes passing electrical current through the metallic article to be cleaned while it is positioned in an aqueous alkali metal hydroxide containing electrolyte to which has been added a cleaning agent selected from the group consisting of 1-hydroxyethylidene-1,1-diphosphonic acid, an alkali metal salt of 1-hydroxyethylidene-1,1-diphosphonic acid and mixtures thereof.

Galling: it is a form of surface damage, arising between sliding solids. It is distinguished by microscopic, usually localized, roughening and creation of protrusions (i.e., lumps) above the original surface. In other words, galling is material transfer from one surface to another caused by movement and plastic deformation.

Hardfacing: a metalworking process where harder or tougher material is applied to a base metal. It is welded to the base material, and it generally takes the form of specialized electrodes for arc welding or filler rod for oxyacetylene and TIG welding.

Inconel: is a registered trademark of Special Metals Corporation that refers to a family of austenitic nickel-chromium-based superalloys. It is primarily composed of nickel (up to 72wt%), chromium (up to 23wt%), iron and molybdenum (up to 10wt% each), niobium and some other trace elements. Inconel alloys are typically used in high temperature applications.

Monel: is a trademark of Special Metals Corporation for a series of nickel alloys, primarily composed of nickel (up to 67wt%) and copper, with some iron and other trace elements.

Nitriding: High temperature gas nitriding (HTGN) or plasma nitriding are diffusion methods, and consist of annealing a stainless steel in N₂-containing gas atmosphere under certain conditions. Atomic nitrogen is absorbed at the surface of the steel and diffuses into the subsurface region.

Pickling: is a metal surface treatment used to remove impurities, such as stains, inorganic contaminants, rust or scale, from ferrous metals, copper, and aluminum alloys. A solution called *pickle liquor*, which contains strong acids, is used to remove the surface impurities. It is commonly used to descale or clean steel in various steelmaking processes.

PMZ: partially melted zone

Precipitation hardening: also called **age hardening**, is a heat treatment technique used to strengthen malleable materials, including most structural alloys of aluminium, magnesium, nickel and titanium, and some stainless steels. It relies on changes in solid solubility with temperature to produce fine particles of an impurity phase, which impede the movement of dislocations, or defects in a crystal's lattice. Since dislocations are often the dominant carriers of plasticity, this serves to harden the material.

Quenching: refers to rapidly cooling a metal. It is most commonly used to harden steel by introducing martensite, in which case the steel must be rapidly cooled through its eutectoid point, the temperature at which austenite becomes unstable. In steel alloyed with metals such as nickel and manganese, the eutectoid temperature becomes much lower, but the kinetic barriers to phase transformation remain the same. This allows quenching to start at a lower temperature, making the process much easier. High speed steel also has added tungsten, which serves to raise kinetic barriers and give the illusion that the material has been cooled more rapidly than it really has. Even cooling such alloys slowly in air has most of the desired effects of quenching.

Scaling: the formation of scales on the metal's surface

Sensitization: the precipitation of chromium carbides at grain boundaries, caused by the high temperatures involved in welding.

Solution heat treatment: a process in which an alloy or metal is heated to a suitable temperature, is held at that temperature long enough to allow a certain constituent to enter into solid solution, and is then cooled rapidly to hold that constituent in solution. Most solution heat treatments soften or anneal.

Spalling: detaching of fragments from a metal's surface

Tempering: is a heat treatment technique for metals, alloys and glass. In steels, tempering is done to "toughen" the metal by transforming brittle martensite or bainite into a combination of ferrite and cementite. Precipitation hardening alloys, like many grades of aluminum and superalloys, are tempered to precipitate intermetallic particles which strengthen the metal. Tempering is accomplished by a controlled reheating of the work piece to a temperature below its lower critical temperature.

Vacuum-Oxygen Decarburization: a process used in the making of stainless steels, like the AOD except no argon carrier gas is necessary. Inside the vacuum chamber, an oxygen lance is blown on top of the melt while argon is bubbled up from the bottom. Desulphurizing slag treatments can be added after the degassing.

VOD: vacuum-oxygen decarburization

YAG: Yttrium aluminium garnet ($\text{Y}_3\text{Al}_5\text{O}_{12}$)

Yttrium aluminium garnet: a synthetic crystalline material of the garnet group. It is also one of three phases of the yttria-aluminium composite, the other two being yttrium aluminium monoclinic (YAM) and yttrium aluminium perovskite (YAP). YAG is commonly used as a host material in various solid-state lasers. Rare earth elements such as neodymium and erbium can be doped into YAG as active laser ions, yielding Nd:YAG and Er:YAG lasers, respectively. Cerium-doped YAG (Ce:YAG) is used as a phosphor in cathode ray tubes and white light-emitting diodes, and as a scintillator.

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