CHAPTER 2

The welding of 12% chromium ferritic steels

2.1 Introduction

The majority of structural steels used today require some or other form of protection against corrosion[1], such as painting or galvanizing. In order to address the need for a more corrosion resistant structural steel, Columbus Stainless has been producing a duplex ferritic/martensitic steel (3CR12) for a number of years. As shown in table 2.1, the steel contains 11% to 12% chromium, and consequently exhibits relatively good corrosion resistance over a wide range of atmospheric and aqueous conditions.

Table 2.1: The chemical composition of 3CR12. (Percentage by weight, balance Fe)

C	Ni	Mn	Si	P	S	Cr
0.03 max	1.50 max	1.50 max	1.00 max	0.03 max	0.03 max	11.0 – 12.0

3CR12 is not a stainless steel, but the rate of metal loss due to corrosion is very low. As a result, the need for expensive coatings in structural applications is diminished[1]. Typically, 3CR12 finds application in motor vehicles, mine ore carts and chutes handling wet farm produce[1].

An apparent limitation to the general application of 3CR12 is its weldability; particularly the reduced toughness in the high temperature heat affected zone adjacent to the weld. This loss in toughness is caused by excessive delta ferrite grain growth in this area[2].

A number of studies have shown that acceptable impact strengths in the welded joint can be obtained when 3CR12 is welded using an austenitic filler metal such as 309 or 309L[3,4]. However, ferrite grain growth in the heat affected zone cannot be prevented and this still presents problems when plate thicker than 3mm is welded. It is not critical for general thin plate applications, and ideally 3CR12 could be used in applications reserved traditionally for carbon steels, such as AISI 1010 used in the transport industry[5].

The heat affected zone properties of 3CR12 welds may be improved if grain growth in the high temperature heat affected zone can be prevented or at least restricted. Ferrite grain growth can be reduced by balancing the chemical composition of the parent metal to include more austenite forming elements[3], but the cost of these additions would increase the price of the steel. The aim of this study was to investigate the potential for raising the local interstitial carbon and/or nitrogen content of the high temperature heat affected zone by diffusion over the fusion line, and thereby stabilizing the austenite (γ) phase locally. The increased stability of the austenite should give rise to a dual phase structure (consisting of ferrite and austenite) at high temperature, and may restrict δ-ferrite grain growth. This chapter discusses current practices of welding 3CR12, and the typical heat affected zone microstructures that can be expected adjacent to the fusion line.

2.2. The Welding of 3CR12

2.2.1 Welding procedures and heat input

The shielded metal arc welding (SMAW) and the gas metal arc welding (GMAW) processes are used in practice to weld thicker sections of 3CR12 (thicker than 3mm)[1]. Plate sections thinner than 3mm are usually welded with gas tungsten arc welding (GTAW)[1]. Submerged arc welding can be used only after extensive procedure testing, as the high heat input and slow cooling rates associated with this process can cause loss of toughness in the heat affected zone[1] due to excessive ferrite grain growth. The heat input should be kept to the minimum to improve the integrity of the weld[1]. During SMAW a short arc length should be maintained, weaving and back stepping should be avoided to minimize the heat input, and the current should be set at the low to medium end of the range recommended by the manufacturer of the welding electrodes[1]. An austenitic filler metal with a low carbon content (such as E309L) is usually recommended[1].

2.2.2. The Heat Affected Zone in welded 3CR12 plate

According to Grobler[3], the heat affected zone in 3CR12 can be divided into three separate zones. The three zones are defined in the following way:

(a) The high temperature heat affected zone (HTHAZ)

The HTHAZ is a narrow coarse grained zone directly adjacent to the fusion line. It is heated into the δ-ferrite phase field during welding (see figure 2.1 and 2.2), and due to the very high peak temperatures reached, excessive ferrite grain growth occurs. The final structure generally consists of δ-ferrite with blocky Widmanstätten martensite on the grain boundaries (see figure 2.3). The fraction martensite depends not only on the relative amount of ferrite and austenite stabilizers present in the steel, but also on the rate of cooling through the austenite (γ) phase field.

Typical peak temperatures reached during welding are shown graphically in figure 2.2. The curves in figure 2.2 were obtained from calculations based on the Rosenthal equation[7]. The Rosenthal equation, together with the relevant calculations, will be discussed in more detail in Chapter 4. The graphs were obtained on the assumption that the fusion line peak temperature is equal to the liquidus temperature of 3CR12. The liquidus temperature was estimated from the phase diagram (figure 2.1) to be approximately 1510 °C. The distance from the center of the heat source to the fusion line can then be measured, and the temperature sequence for a point at this distance from the fusion line can be calculated. The results indicate that the large delta ferrite grains form at a temperature above a certain peak temperature at a specific distance away from the fusion line. The width of the heat affected zone can be observed and measured by applying a light etch with Kalling's no. 2 reagent, and was found to be 0.5mm for the conditions in figure 2.2.

Grain boundary austenite formed during the heating cycle effectively inhibits grain growth, and transforms to martensite on cooling. During cooling, some of the δ-ferrite transforms to grain boundary austenite. In the case of a HTHAZ with no austenite on the ferrite grain boundaries, the composition of the heat lies to the extreme right of the dual phase austenite and ferrite region in the iron-chromium phase diagram (figure 2.1). As a result the time available for austenite formation is limited during the heating and cooling cycles[4]. The width of the HTHAZ depends on the heat input and the chemistry of the specific heat, and varies between 0,1 mm and 0,5 mm[4]. Thermal cycle predictions with the Rosenthal equation can be used to predict the width of the HTHAZ. (See chapter 4).

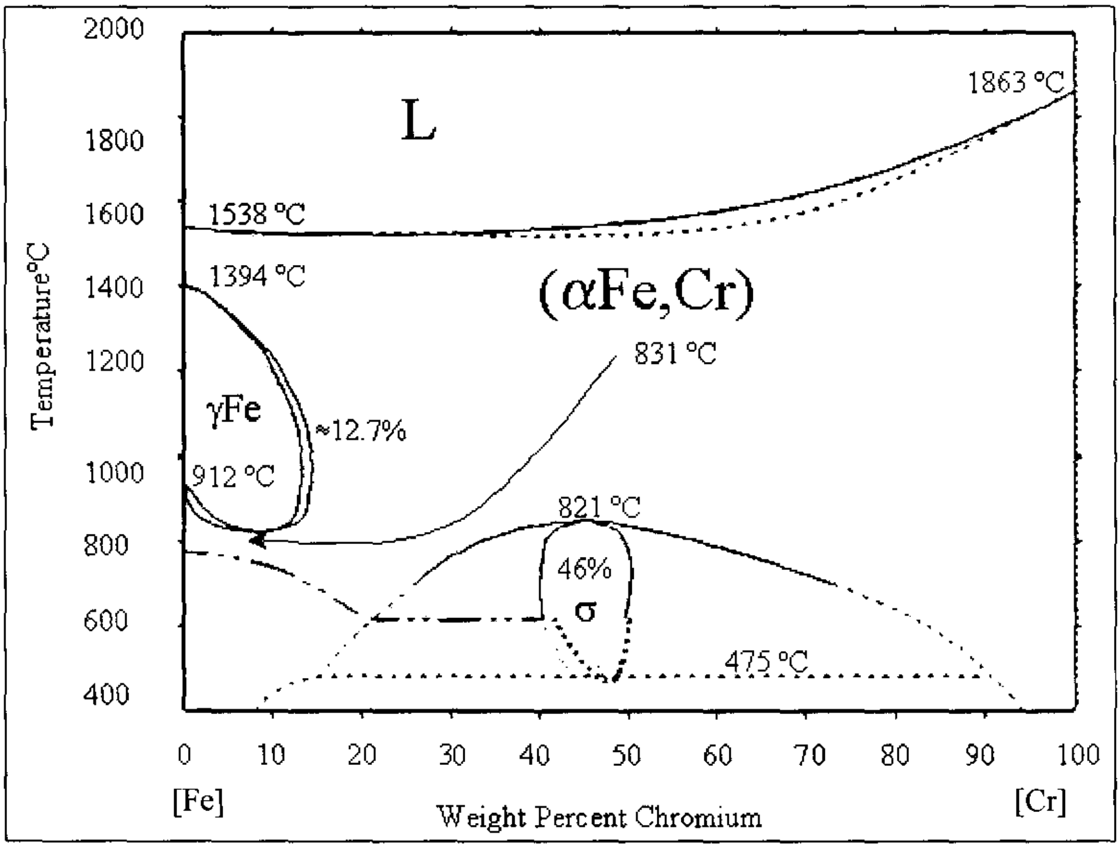

Figure 2.1: The iron-chromium equilibrium diagram

b) The duplex zone

A much wider duplex zone is located adjacent to the HTHAZ. This zone consists of fine grained ferrite and untempered intergranular and transgranular martensite (figure 2.4). This zone originated when the steel was heated into the dual phase (austenite + ferrite) phase field in figure 2.1. The phase composition of the duplex zone varies across the width of the zone. As discussed earlier the high temperature heat affected zone adjacent to the fusion line consists of large delta ferrite grains with blocky martensite on the grain boundaries. Within the duplex zone the fraction martensite increases with distance from the fusion line, while the delta ferrite grains become smaller. The ferrite grains tend to be elongated further away from the fusion line. The width of the duplex zone is dependent on both the individual heat composition and the heat input during welding. This zone exhibits a high hardness, mainly because of the high fraction of untempered martensite present in the microstructure

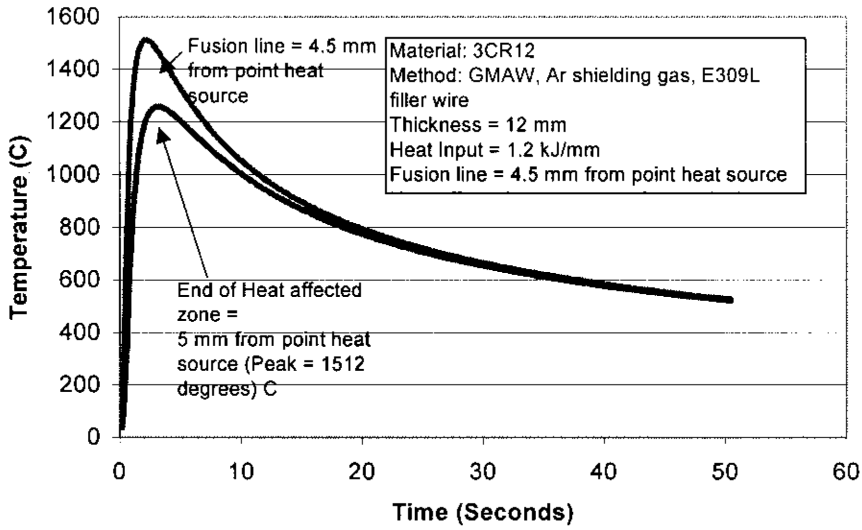

Figure 2.2: The welding temperature sequence from Rosenthal[7].

. (c) The low temperature heat affected zone.

The low temperature heat affected zone is more distant from the fusion line, and was heated to below the transition temperature to the dual phase (austenite + ferrite) region (see figure 2.1) during welding. The hardness of the parent plate remains unchanged in this zone[3]. The steel exhibits a relatively high tempering resistance, and the time at high temperature is too short for any additional tempering to occur[6]. The low temperature heat affected zone does not exhibit significant metallurgical changes during welding[3].

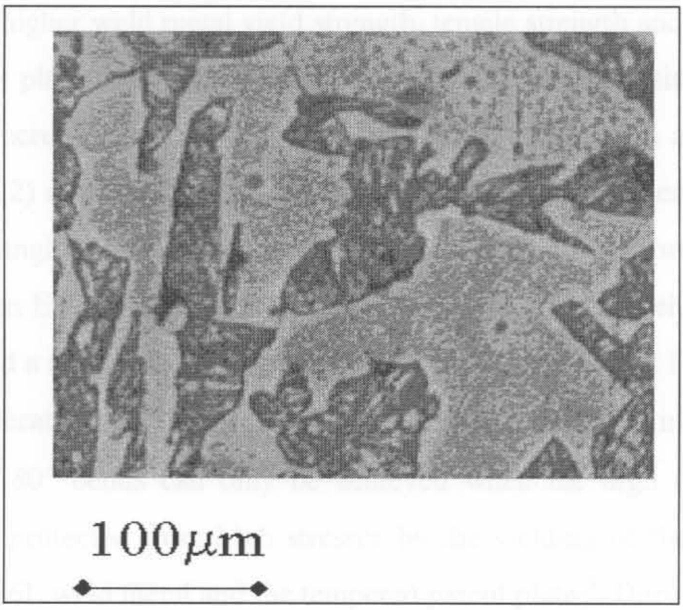

Figure 2.3: The high temperature heat affected zone.(GMAW, HI=1.3kJ/mm, etched by quick swabbing with Kalling's reagent nr.2[8]).

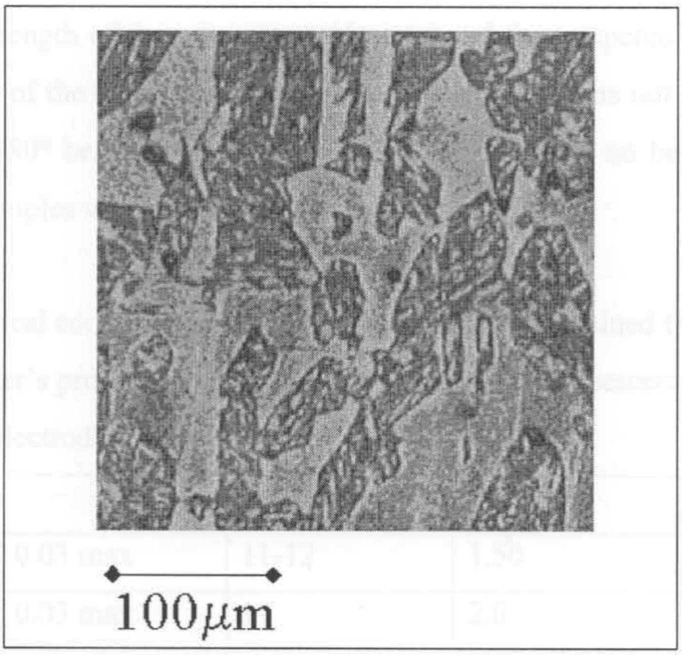

Figure 2.4: The duplex zone. (GMAW, HI = 1.3 kJ/mm, etched by quick swabbing with Kalling's

2.2.3 Filler Metals for welding 3CR12

Generally, the susceptibility to brittle fracture in the HTHAZ in 3CR12 welds increases with a higher weld metal yield strength, tensile strength and work hardening rate[3]. In bead on plate bend tests performed by Grobler[3], the maximum bend angle decreased with increasing weld metal hardness. An E3CR12 (with a composition as shown in table 2.2) electrode yielded the highest weld metal hardness, but the lowest maximum bend angle. The bend angle obtained from the E3CR12 welded plates was only 13° . With an E316L electrode (see table 2.2) a much lower weld metal hardness was obtained, and a maximum bend angle of 180° could be reached. Fracture occurred in the high temperature heat affected zone adjacent to the fusion line. This suggests that successful 180° bends can only be achieved when the high temperature heat affected zone is protected from high stresses by the yielding of the adjacent softer material (the E316L weld metal and the tempered parent plate)[3]. During bending, most of the deformation in the specimens welded with the E316L occurred in the low strength low temperature heat affected zone, with some deformation in the weld metal, and almost none in the high temperature coarse grained zone[3]. As a result of the relatively low strength of the adjacent weld metal and the tempered parent plate, the fracture strength of the high temperature heat affected zone was not exceeded during bending[3], and 180° bends could be achieved. Unfortunately, no bending tests were carried out on samples welded with E309L electrodes.

Table 2.2: Chemical composition of different electrodes as obtained from the supplier's product information. All values are mass percentages. An E307 electrode was not tested by Grobler.

Electrode	C	Cr	Mn	Ni
E3CR12	0.03 max	11-12	1.50	1.50
E316L	0.03 max	17	2.0	12
E309L	0.03 max	23	2.0	13.5
E307	0.16	15-17	2.0	12-13

As a rule, a 309L filler metal is recommended when welding 3CR12 for high integrity structural applications[1]. This austenitic type filler metal has a low yield strength relative to the 3CR12 parent plate, and consequently protects the grain growth zone against brittle fracture by absorbing a high fraction of the plastic deformation during bending[3]. At high dilutions (higher than 43% [3]) weld metal martensite forms. The yield strength of the weld metal will increase due to the presence of weld metal martensite, and this would be detrimental to the fracture toughness of the weld as the weld metal would not be able to absorb sufficient plastic deformation. As a result, the fracture strength of the high temperature heat affected zone may be exceeded during bending. The major disadvantages of the E309L filler metal are that it is relatively expensive and that the possibility of galvanic corrosion in certain environments exists[3]. In order to avoid galvanic corrosion and provide an electrochemical match between the weld and the parent plate, the composition of the weld has to closely resemble that of the parent plate. Such a matching electrode, for example E3CR12, would lie in the ferrite-martensite phase field on the Schaeffler diagram (figure 2.5), indicating that untempered martensite would be present in the weld metal. As a result, E3CR12 welds have high yield strengths and cannot protect the grain growth zone from brittle fracture by plastic deformation of the weld metal[3]. For this reason, the E3CR12 filler metal is not recommended for welding of 3CR12 in any structural type application, particularly not if the impact strength of the welded joint is critical[3].

The conclusion can be drawn that neither the E309L, nor the E3CR12 filler metals, meet both the electrochemical and the mechanical requirements as filler metal for welding 3CR12[3]. In practice, the particular application would determine the choice between these two filler metals[3], as well as between other filler metals such as E316L (See the Schaeffler diagram in figure 2.5).

2.3 Summary

3CR12 offers good corrosion resistance in mild environments. During the welding of thick sections of 3CR12, embrittlement of the high temperature heat affected zone may occur as a result of δ-ferrite grain growth in the high temperature heat affected zone adjacent to the weld. The detrimental effect of the ferrite grain growth may be reduced by welding with a relatively soft, low carbon austenitic filler metal such as E309L or E316L. Plastic deformation in the weld metal and in the parent metal protects the heat affected zone by preventing the fracture stress of the high temperature heat affected zone from being exceeded. However, the chemical compositions of E309L and E316L are notably different from that of 3CR12, and galvanic corrosion may occur if these filler metals are used in highly corrosive environments. An E3CR12 filler metal provides the electrochemical match to the parent metal to prevent galvanic corrosion, but as a result of the large quantity of untempered martensite in the weld metal, the weld will be very hard. As a result, the E3CR12 weld metal would be unable to absorb sufficient plastic deformation to protect the heat affected zone from stresses that may exceed the fracture stress of the zone.

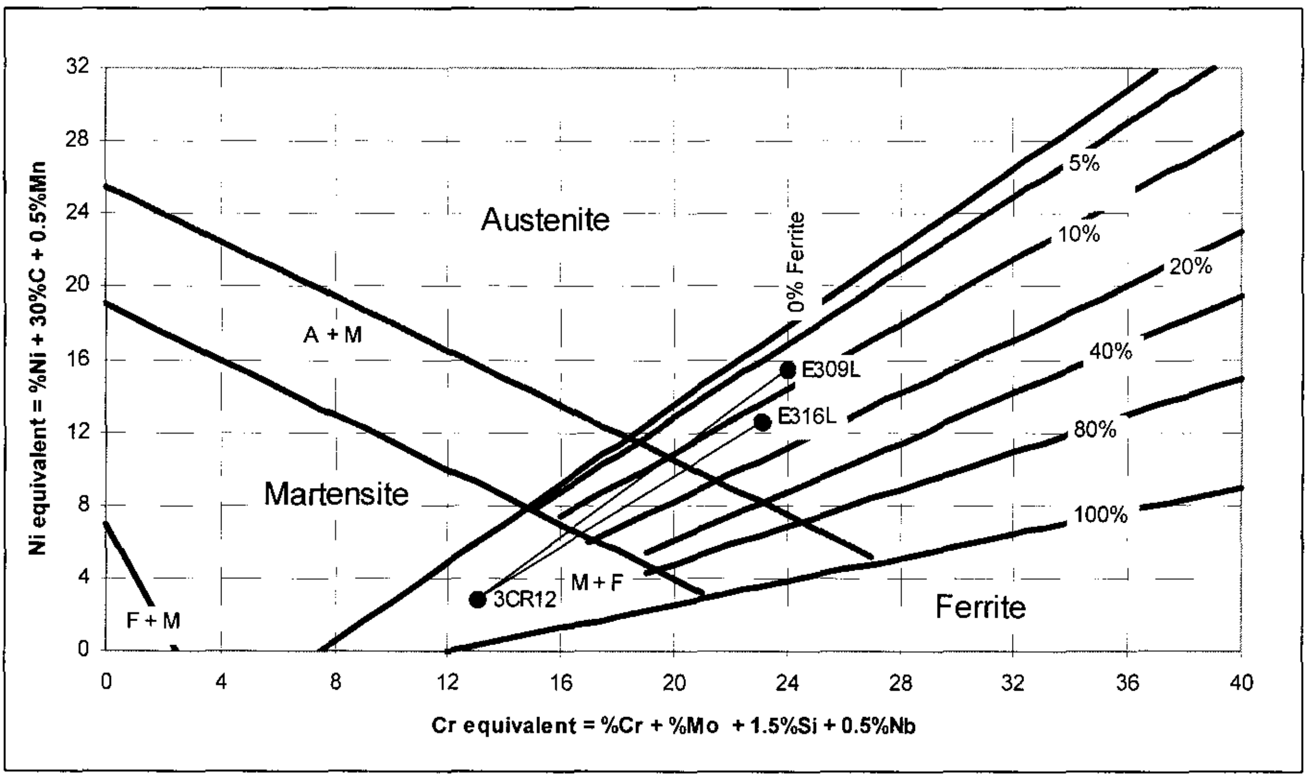

Figure 2.5: The Schaeffler diagram, showing different filler metal compositions[2].

The conclusion can be drawn that the welding of 3CR12 is a compromise between the corrosion resistance and the impact strength of the welded joint. The relative importance of the two characteristics in a specific application would determine the choice of the filler metal.

2.4 References

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