

THE EMPHASIS OF PHASE TRANSFORMATIONS AND ALLOYING CONSTITUENTS ON HOT CRACKING SUSCEPTIBILITY OF TYPE 304L AND 316L STAINLESS STEEL WELDS

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Abstract:

Hot cracking is a significant problem due to transformation of retained ferrite into sigma phase, which results preferential corrosion of ferrite. The Hot Cracking Susceptibility is high for fully austenitic compositions but specimens with 5 to 30% ferrite were quite resistant to cracking. Hot cracking in 304L and 316L is amplified by low-melting eutectics containing impurities such as S, P, Si, N. It could be diminished by small increase in C, N, Cr, Ni, Si or by substantial increase in Mn content. To avoid formation of excess ferrite and to control cracking, weld metal should have a ferrite content of at least 3 to 5 ferrite number, the present investigation evaluates the hot crack susceptibility of 304L and 316L type austenitic stainless steels to the chemical composition of the alloying material. The propensity for hot cracking is determined primarily by the hot cracking susceptibility factor. To determine the suitable composition and the corresponding ferrite numbers for type 304L and 316L the Schaeffler-DeLong diagram is used.

Keywords: Austenite, Ferrite, Hot Cracking Sensitivity Factor, Stainless Steel, Phase Diagrams.

1. Introduction

American Iron and Steel Institute (AISI) type 304 is the foundation of 300 series austenitic stainless steels is by far the oldest, largest and most important group in the stainless steel range [1]. While type 316L substitutes approximately 2% Mo for a nearly equal amount of Cr to improve pitting corrosion resistance. The L grades represent low-carbon variants with a nominal carbon level of 0.03 % depending on the specification to which it is manufactured, which are used for general stainless steel fabrications, elevated temperature applications and resistance to pitting corrosion respectively [2]. 300 series of austenitic stainless steels usually solidify during welding as a mixture of austenite and ferrite. The ferrite almost fully transforms to austenite on cooling, but there could be retention of a few percent of δ ferrite in the weld metal which leads to hot cracking [3].

Hot cracking refers to cracking that occurs during welding, casting, or hot working at temperatures close to the melting point of the material [4,5]. Several theories had been advanced to explain the phenomenon of solidification cracking [6,7,8,9]. The initial theories took into account that hot cracking is associated with segregation [6,7]. Later, Scherer et al claimed that crack-resistant weld deposits can be produced if ferrite is restricted to 5–35% in the completed weld [10]. Maziasz stated that irradiation stability and creep properties can be improved by surplus addition of elements like phosphorus and boron and a minimum content of phosphorus [11]. However, Hull examined that impurity elements such as sulphur and phosphorus, and minor alloy elements such as boron, silicon, titanium and niobium promote hot cracking, particularly in fully austenitic steels [12]. Savage et al and Lundin et al had also explained the Varesstraint test and its modifications that had been used in a large number of studies on hot cracking susceptibility of materials, for weldability comparisons and process and alloy development. [13,14]. However, these theories were not entirely satisfactory, as several

exceptions could be found and the freezing range appeared to be only one of many factors influencing cracking. Finally Borland proposed the ‘Generalized Theory’ of cracking, which was further modified by Smith, Matsuda and Clyne [15,16,17].

The purpose of this review is to examine the current understanding of Hot Cracking Susceptibility (HCS) in type 304L and 316L austenitic stainless steels with particular emphasis on significant impurities and alloying elements along with influence of phase transformations on hot cracking.

2. The theory of Hot Cracking

According to the distribution of the liquid and solid phases for a binary alloy, Borland stated that solidification involves four stages [8,9]. In stage 1, there is relative movement between continuous liquid and dispersed solid phase, in subsequent stage 2, due to interlocking between solid dendrites only the liquid is capable of movement, the liquid can heal any cracks formed. In stage 3, the solid crystals are in an advanced stage of development and the free means of access of liquid is averted. Cracking can occur, which can not be filled by the remaining liquid phase due to little quantity of liquid. If a stress is applied which exceeds the tolerance of the material, this stage during which much of the cracking occurs, is called the critical solidification range (CSR). In final stage solidification is complete and there is no possibility of cracking.

Borland stated that for high cracking susceptibility, in addition to a wide freezing range, the liquid must be distributed as films or as isolated droplets that permits high stresses to be built up between grains on the basis of actual behavior depending on the interfacial energies of the solid and liquid phases. Clyne et al. found few practical difficulties in Borland’s theory [16]. Later, Matsuda had proposed a modified theory that suggested that crack initiation and propagation should be considered separately as shown in figure 3 [17]. He stated that the critical solidification range (CSR) starts at a higher temperature near to the liquidus and further divided into two stages, an initiation stage (stage 3(h)) and a propagation stage (stage 3(l)) as shown in figure 1.

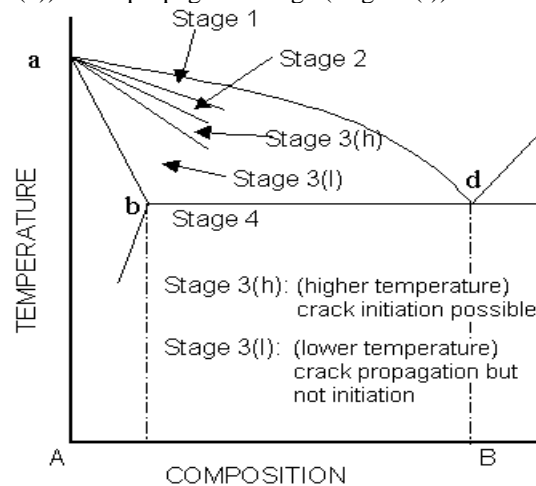


Fig. 1 Modified concept of the effect of constitution on cracking susceptibility in binary systems [Ref.17]

3. Solidification-Phase relationship Diagrams

3.1 Solidification-Phase relationships in the Fe–Cr system

There is complete solubility of Cr in iron and solidification of all Fe-Cr alloys occurs as ferrite at elevated temperatures stated Sedriks [18]. Sindo Kou explained at low chromium concentrations a “gamma loop” of austenite exists in the temperature range 912 to 1394°C. 304L and 316L steels, as having more than 16wt% Cr will be fully ferritic at elevated temperatures, while alloys with less Cr (<12.7%) will form at least some austenite within the gamma loop as indicated in figure 2 [19]. In the Fe-Cr diagram for 304L and 316L steels due to high Cr (if Cr>20%) availability in both steels, a hard, brittle, low-temperature equilibrium phase, sigma phase (σ), is present, with (Fe, Cr) stoichiometry. The kinetics, of formation of sigma, is quite sluggish and precipitation requires extended time in the temperature range 600 to 800°C. A dotted horizontal line can be seen within the $\sigma + \alpha$ phase field at the temperature of 475°C, designates a phenomenon called 475°C embrittlement, resulting from the formation of coherent Cr-rich precipitates within the α -matrix. These precipitates are known as α' (alpha prime) and form within the temperature range 400°C to 540°C. The presence of the above precipitates results in severe embrittlement in 304L and 316L steel [17].

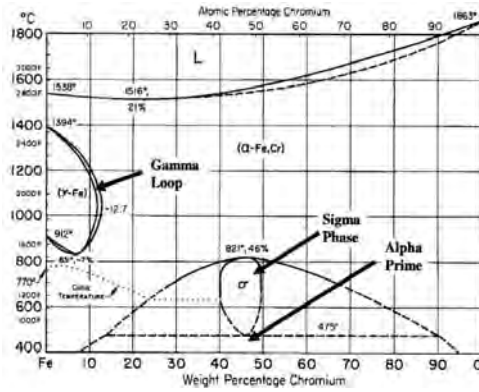


Fig. 2 The Fe-Cr phase diagram [Ref. 19]

3.2 Solidification-Phase relationships in the Fe–Cr–C system

Carbon is an austenite-promoter and its addition in Fe- Cr system will expand the gamma loop. In order to view the Fe-Cr-C ternary system, Lippold et al. constructed a pseudobinary diagram (isopleths) which represents a two-dimensional projection of a three-dimensional diagram [20] as in figure 3, a pseudo-binary diagram, based on 17 wt% Cr with variable carbon content as a function of temperature. Here C_1 is a $(Cr,Fe)_{23}C_6$ carbide and C_2 is a $(Cr,Fe)_7C_3$ carbide, is a result of the addition of carbon in the alloy.

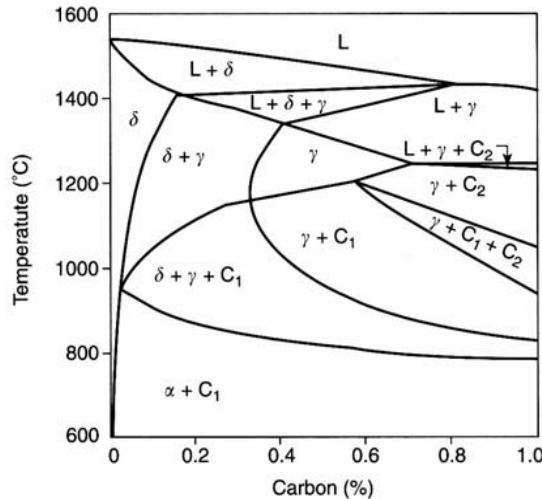


Fig. 3 Fe-Cr-C pseudo-binary diagrams at 17% Cr [Ref. 20]

3.3 Solidification-Phase relationships in the Fe–Cr–Ni system

Folkhard explained that the liquidus projection starts at the peritectic reaction on the Fe–Ni system ($\delta + L \leftrightarrow \gamma$) and moves down to the eutectic reaction ($L \leftrightarrow \gamma + \delta$) on the Cr–Ni system in the Fe–Cr–Ni ternary system [22]. On the Ni and Cr enriched side a ternary eutectic is formed at 1300°C. Most stainless steel compositions in wide use occur on the iron-rich side of the ternary between 50 and 70 % iron. Liquidus and solidus projections of the Fe–Cr–Ni system (70% Fe) shown along with the constituent binaries is shown in figure 4.

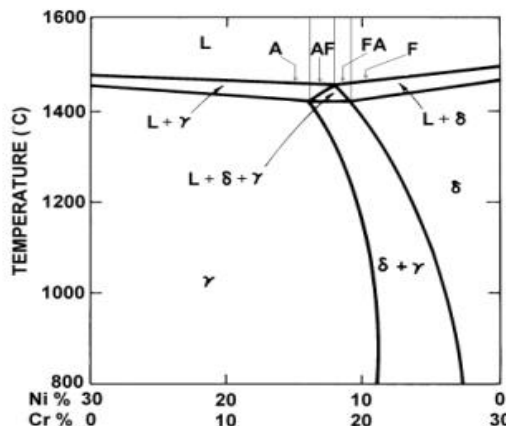


Fig 4. Pseudobinary section of the Fe–Cr–Ni ternary diagram at 70% Fe, showing solidification modes [Ref. 17]

Four distinct modes are normally considered, viz., A, AF, FA, F. Alloys solidifying in the A mode will remain unchanged to low temperatures, while those solidifying as AF would form some eutectic ferrite. Compositions that solidify in the FA and F modes pass through the ($\delta \leftrightarrow \gamma$) two phase region and may re-enter the single-phase austenite field. This is due to the asymmetry of the two-phase field towards the primary ferritic side of the diagram, as seen in figure 5. Thus, alloys such as Type 304 and 316 that are fully austenitic at room temperature enter this two-phase region after AF/FA solidification and may undergo solid-state transformation to a fully austenitic structure. Fredriksson had researched that the transition from peritectic to eutectic reaction in Fe–Cr–Ni alloys occurs at 17.2% Cr and 11.9% Ni. Under equilibrium conditions, the peritectic reaction occurs only at iron contents above 75 % [23].

David et al explained that solidification in leaner alloys such as Type 304, 316 tends to be peritectic while higher alloyed grades undergo eutectic type of reaction. For most stainless steels this means that the composition should be controlled to achieve the FA solidification mode, resulting in the presence of FN 3-20 in the weld deposit[24]. Solidification mode was found major determinant of hot cracking susceptibility, ensuring an FA or F mode ensures the best resistance to hot cracking. Details of various solidification modes with their equivalent ratio, reaction and microstructure are presented in table 1.

Table 1 Solidification Modes, Reactions and Final Microstructure Creq/Nieq ratio and solidification mode association [Ref. 17]

Solidification Mode	Equivalent ratio	Reaction	Microstructure
A Fully Austenitic	$Cr_{eq}/Ni_{eq} < 1.25$	$L \rightarrow L + A \rightarrow A$	Fully Austenitic
AF Austenitic Ferritic	$1.25 < Cr_{eq}/Ni_{eq} < 1.48$	$L \rightarrow L + A \rightarrow L + A + (A + F)_{eut} \rightarrow A + F_{eut}$	Ferrite at cell and dendrite boundaries
FA Ferritic Austenitic	$1.48 < Cr_{eq}/Ni_{eq} < 1.95$	$L \rightarrow L + F \rightarrow L + F + (F + A)_{per/eut} \rightarrow F + A$	Skeletal and/or lathy ferrite resulting from ferrite to austenite transformation
F Ferritic	$1.95 < Cr_{eq}/Ni_{eq}$	$L \rightarrow L + F \rightarrow F \rightarrow F + A$	Acicular ferrite or ferrite matrix with grain boundary austenite and Windmastätten side plates

4. Types of Cracking in 304L and 316 L steel welds

4.1 Solidification cracking

All metals show a typical dependence between ductility and temperature, increasing temperature in the solid state up to the solidus temperature (T_s) leads to a higher ductility. But above T_s a sharp decrease in ductility due to the presence of a second, liquid phase occurs, called the Brittle Temperature Range (BTR). If stresses are applied within the BTR, solidification cracking can take place. To minimize the BTR and to prevent solidification cracking the presence of elements, which can form phases with low melting points should be avoided. During cooling from the liquid weld pool these critical phases remain as a liquid at the solidification front. Additional stresses, caused by the inhomogeneous welding temperature field can lead to cracking [17, 21].

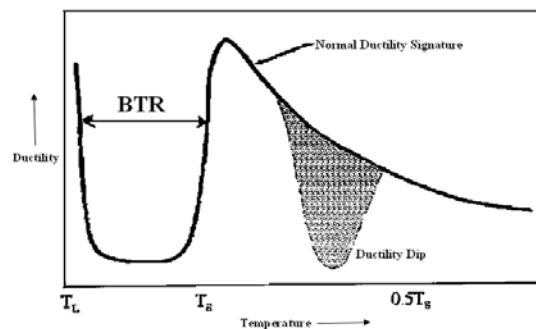


Fig 5 Appearance of BTR and Ductility Dip [Ref. 39]

4.2 Liquification cracking

Liquification cracking can occur, if hot cracking sensitive materials are heated up to temperatures above the melting point of the low melting critical phases but below the “main” solidification temperature of the alloy [17,39]. Also in this case, stresses can lead to cracking. Control of impurity levels and minimizing heat input can reduce or eliminate this form of cracking. Minimal ferrite number to control liquification cracking for 304L stainless steel is 4.0 and for 316L is 2.5 [38].

4.3 Ductility dip cracking

High chromium nickel-base alloys show in a temperature range solid state grain boundary embrittlement phenomenon which is called Ductility Dip Cracking (DDC) as displayed in figure 5 [39]. Fine particles at the grain boundaries, especially carbides of type $M_{23}C_6$ can prevent grain boundary gliding and in further consequence the formation of voids and small cracks along grain boundaries.

5. The Emphasis of δ Ferrite on hot cracking

By assuring solidification as primary ferrite, the potential for cracking is effectively eliminated. The presence of delta ferrite causes a number of problems with the austenitic stainless steels:

1. The cracking sensitivity gain increased as δ ferrite content is increasing and which has inverse effect on, the ductility because of martensite formation and thus the potential for fracture increases [12,25].
2. If secondary operations like Electropolishing are performed, the delta ferrite will preferentially dissolve leaving a white or dull surface, especially true for welded parts [40].
3. As the δ ferrite is usually controlled to prevent microcracks as well as refines the grain size of the solidified metal, which results in better mechanical properties and cracking resistance in stainless steel welds [19,35].
4. Ductility of ferrite at high temperatures is greater than that of austenite, allowing relaxation of thermal stresses. The lower thermal expansion coefficient of ferrite as compared to austenite results in less contraction stresses and fissuring tendency [8].
5. Volume contraction associated with the ferrite-austenite transformation reduces tensile stresses close to the crack tip, which decreases cracking tendency [30].
6. Coarse grain formation in the HAZ occurring by recrystallisation and grain growth in fully austenitic metals increases susceptibility to liquation cracking, while ferrite forming compositions are not susceptible [29].

As determined by the Welding Research Council, the optimum δ ferrite content for welding is 2 – 5% i.e. to avoid cracking; weld metal should have a ferrite content of atleast 3 to 5 ferrite number [40]. To determine the suitable composition and the corresponding ferrite number for type 304L and 316L, the Schaeffler-DeLong Diagram, Figures 6A and 6B are used. Figure 6 A has the ASTM composition limits for Type 304L and Figure 6 B the limits for Type 316L superimposed with actual ferrite numbers from 172 heats [26]. The black dots represent actual heats. For 304L, the group clustered around 5 – 6% is older heats, the group around 10-12 FN represent the newer heats. Heat treatment, which involves holding at 1040 - 1100°C for at least will dissolve δ

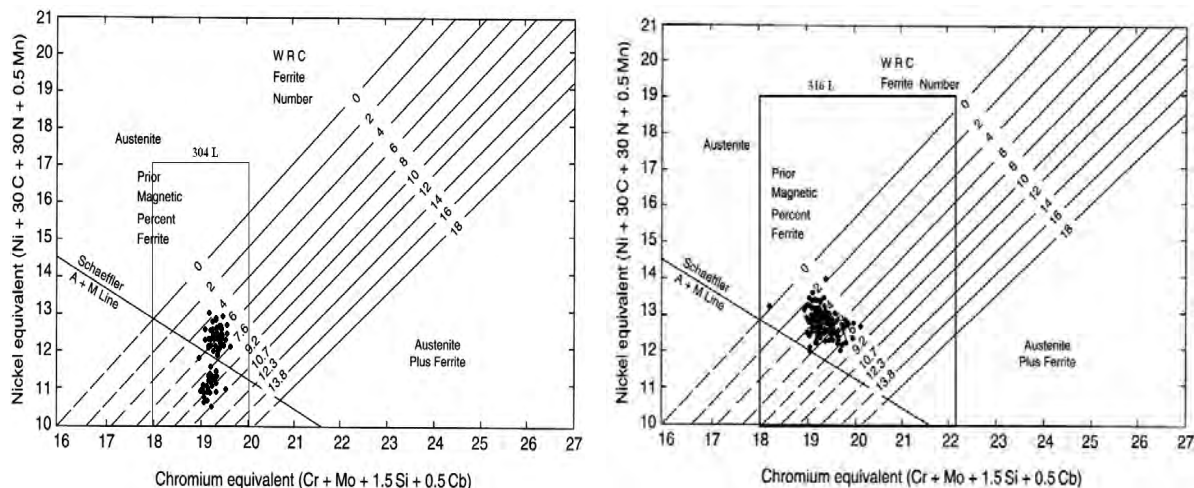


Fig. 6 A and 6 B The, Schaeffler-DeLong Diagram for determining the δ ferrite with the composition limits for 304L, 316L [Ref. 40]

ferrite up to 7-8%. Above that level, some δ ferrite remains in the structure as a second phase. Additional heat treatment or longer time at temperature has little effect on reducing the ferrite effect when the content is this high. For 316L, most of the 72 compositions are clustered around 4% FN and later heats around 6 FN [40].

6. Role of Alloying Elements

6.1 Chromium

The primary objective of the chromium addition is to provide corrosion protection to the steel. In the 304L and 316L austenitic stainless steels it is normally present in the range 16-22 wt% [2].

1. Chromium is also a key ingredient in the formation of intermetallic compounds as α (alpha prime) which causes embrittlement in 304L and 316L steel. The percentage of chromium increases the stability of the oxide since it has a much higher affinity for oxygen than does iron [9].
2. Chromium is a substitutional element in BCC and FCC, and so from a mechanical properties standpoint, chromium will provide some degree of solid solution strengthening [17].
3. Chromium is also known as strong carbide former as $(Cr,Fe)_{23}C_6$ carbide, $(Cr,Fe)_7C_3$ carbide and the most common Cr-rich carbide is the $M_{23}C_6$, where "M" is predominantly chromium but may also have some fraction of Fe and Mo present [21].

6.2 Nickel

The prime purpose of nickel's addition is to encourage sufficient stable austenitic phase field [17]. In the 304L and 316L austenitic stainless steels availability of nickel is in the range of 8-15 wt% [1,2].

1. Nickel is also a good solid solution strengthener with addition of nickel to a $Fe_{20}Cr$ alloy, in an aggressive chlorine containing environment, a decrease in stress corrosion cracking resistance occurs [32]., [9].
2. Presence of high Ni content decreases the crevice corrosion resistance of austenitic stainless steels [33].
3. Fe-Cr-Ni(Mo) alloys system in nickel austenite steels, exhibit a wide range of mechanical properties that are unparalleled by any other alloy system today [34].

6.3 Manganese

Its ability to stabilize the austenite in steel manganese is used the 304L and 316L steels, in which its content varies from 1-2.5 %. Manganese is essential to steel making because of these key properties [39,40].

1. The original purpose of manganese addition was the prevention of solidification cracking associated with the formation manganese sulfide as manganese combines more readily with sulfur, than does iron [34].
2. It has powerful deoxidation capacity, In austenitic stainless steels, such as the 304 type it appears to have little effect in promoting austenite versus ferrite. The effect of manganese on the mechanical properties of the alloy is minimal [17].
3. The only negative aspect of manganese is its tendency to form inclusions in the steel, especially in the presence of sulphur [40].

6.4 Molybdenum

Molybdenum is used in 304L and 316 L stainless steels in amounts 0.213 to 3 wt%. Even such relatively small percentages of molybdenum have powerful effects in improving the resistance to pitting in chloride environments and to crevice in both Fe-Cr alloys and Fe-Cr-Ni alloys.

1. Molybdenum reduces the intensity of the oxidizing effect required to insure passivity and decrease the tendency of previously formed passive films to break down [17].
2. Molybdenum is a ferrite-promoting element and its presence will promote ferrite formation and retention in the microstructure [21].
3. Molybdenum is a carbide former [34].

6.5 Silicon

Silicon, a ferrite stabilizer is added for deoxidation during melting. In 304L and 316L steel, its content ranges from 0.4 to 1.0 % [2]. Silicon segregates during solidification particularly in combination with nickel, resulting in formation of low melting eutectic constituents. For the above reason, silicon is held below 1 wt%.

1. It has been found high addition of silicon in austenitic stainless steels, not only improves resistance to oxidation but also prevents carburising at elevated temperatures [35,40].
2. Silicon forms a number of silicides ($FeSi$, Fe_2Si , Fe_3Si , Fe_5Si_3) and a Cr_3Si intermetallic element, all of which tend to embrittle the structure [35].
3. Austenitic grade 304L and 316L steel can be quite sluggish in the molten state during solidification and the addition of silicon in the weld filler metal can improve their fluidity [34].

6.6 Sulphur

Sulphur is present in 304L and 316L steel as undesirable impurity (0.022 to 0.033 %). It is strongly rejected into the liquid during solidification of austenite, rapidly lowering the melting point of the interdendritic liquid [13]. A base material with extremely low sulphur content may produce a nearly unweldable alloy yet one that is crack insensitive when fusion welded [38].

1. It also forms low-melting sulphide films along the interdendritic and grain-boundary regions [21].
2. Sulphur is almost insoluble in all three major constituents viz. Fe, Cr, Ni of steel. The phase diagrams for sulphur binaries with all three elements show wide and deep solid-liquid regions with low partitions coefficients for sulphur in austenite as shown in table 2[41].
3. In alloys rich in austenite stabilizers, heats with excessive sulphur content may experience hot cracking or may affect the amount of hot cracking as opposed to its quality of weld puddle control or penetration [14].

Table 2, Partition coefficients of elements of 304 L and 316 L steel promoting hot cracking in austenite and ferrite, constitution and melting points of possible low-melting phases. [Ref.22,36]

S.No.	Constituent	Temperature °C	Partition Coefficient		Low Melting Phases	
			δ	γ	Structure	Melting points °C
1	Sulphur	1365	0.091	0.035	Eutectic Fe-FeS	988
					Eutectic Ni-NiS	630
2	Phosphorous	1250	0.23	0.125	Eutectic Fe-Fe ₃ P	1048
					Eutectic Ni-Ni ₃ P	875
3	Silicon	1300	0.77	0.52	Eutectic Fe-Fe ₂ Si	1212
					Eutectic NiSi-	964
					Ni ₃ Si ₂	996
					NiSi γ	

6.7 Phosphorous

Phosphorus ranges from 0.31% to 0.45% in 304L and 316L steels. Like sulphur, phosphorous also forms low-melting eutectics with iron, chromium and nickel. The cracking tendencies of P and S tend to be combined and are assumed additive as shown in figure 8.

1. It particularly harmful in 304L and 316 L since it has a strong tendency to spread as liquid films [5].
2. Low diffusivity of P in austenite and ferrite phases even at high temperatures virtually precludes homogenization [22].
3. Manganese additions are well-known to decrease cracking in high-S steels by forming higher-melting MnS-eutectics in preference to FeS [12].

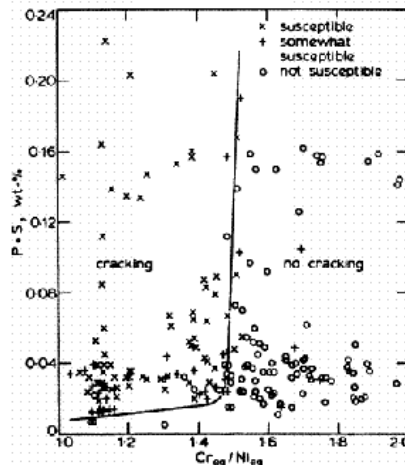


Fig. 7, Cracking susceptibility of 300 series stainless steel based on Cr/Ni equivalent [Ref. 31]

6.8 Nitrogen

Nitrogen has potent effects on the microstructure and hence is expected to have strong influence on hot cracking behaviour of type 300L and 316L stainless steel [44].

1. N appears to act synergistically with S to increase cracking in fully austenitic 316L weld metal; low-S weld metal shows no effect on increasing N levels[17].
2. The beneficial effects of nitrogen to a refinement in the dendritic structure [14].
3. Nitrogen is a strong solid solution strengthening agent and even small additions of N can increase dramatically the strength of austenitic alloys and that is why their content levels should be controlled carefully in order to achieve the desired microstructure balance [45].

7. Welding of 304L and 316 L stainless steel

In general, 304L and 316L steels are readily weldable with common arc processes and other techniques, but very hard to keep flat, as the coefficient of linear expansion is 1.7 times that of mild steel. 308L may be used as filler for 304L, as 304L being low carbon versions suffer from weld decay. For arc welding processes, the hot cracking tendency can be predicted by considering the relative amounts of the ferrite stabilizers to the amount of austenite stabilizers. While for fusion welding processes with either very high cooling rates or very low cooling rates the weld nugget morphology becomes more important than the primary mode of weld pool solidification. The beam processes are best suited to applications that favor very high aspect ratio weld nuggets (high depth to width ratio). The best gas for MIG welding for 304L and 316L is 97.5% Argon + 2.5% CO₂. Virtually all standard specifications in current use for the procurement of 304L and 316L allow for the possibility of receiving material overly rich in austenite stabilizers such that weld solidification as primary austenite could occur and lead to hot cracking in a production weld [2,46].

The plot for SA 240 type 304L shows that approximately one tenth of the area enclosed by the limit lines will solidify as austenite, and additional fourth will solidify as primary austenite, one half will solidify as primary ferrite, while the balance will solidify as ferrite. Similarly for SA 240 type 316L, approximately one third of the possible compositions will solidify as austenite, one fourth as primary austenite, one third as primary ferrite and the balance as ferrite. It should be noted that the maximum carbon content permitted by the SA 240 was used for the calculations. In addition, due to the absence of specification limits for nitrogen, a typical amount of 0.07% was assumed [2].

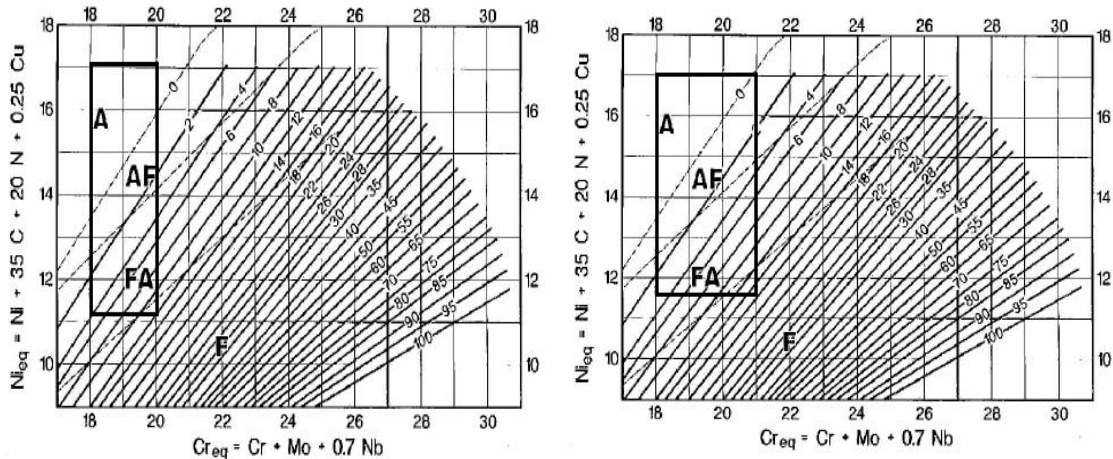


Fig 8 A, 8 B, WRC 1992 constitutional diagram with SA 240 type 304L, SA 240 type 316L stainless steel [Ref.2]

8. Quantification of Hot Cracking Susceptibility in 304L and 316L austenitic stainless steel

Quantification of HAS for various specifications of types 304L and 316L as displayed in table 3 can be done with the help of following equation. But it is advisable to have S<0.035%, Ni<1.0%, Mn>0.8%, C<0.15% and Mn/S>35 [25].

$$\text{Hot Cracking Susceptibility (HCS)} = \frac{(S + P + Si/25 + Ni/100) \times 10^3}{3Mn + Cr + Mo + V} \quad (1)$$

Table.3 Composition ranges for various specifications of types 304L, and 316L [Ref.2]

S.No.	Specification	Type	C*	Mn*	P*	S*	Si*	Cr	Ni	Mo	N*	Fe
1	SA 182 (forging)	304L	0.035	2.00	0.045	0.030	1.00	18.00-22.00	8.00-13.00	N/A	N/A	Balance
2	SA 240 (plate)	304L	0.030	2.00	0.045	0.030	0.75	18.00-20.00	8.00-12.00	N/A	0.10	Balance
3	SA 312 (pipe)	304L	0.035	2.00	0.040	0.030	0.75	18.00-20.00	8.00-13.00	N/A	N/A	Balance
4	Typical SA240	304L	0.024	1.75	0.031	0.022	0.40	18.41	8.75	0.213	0.033	Balance
5	SA 312(pipe)	316L	0.035	2.00	0.04	0.03	0.75	16.0-18.0	10.0-15.0	2.00-3.00	N/A	Balance
6	SA 240(plate)	316L	0.03	2.00	0.045	0.03	0.75	16.00-18.00	10.00-14.00	2.00-3.00	0.1	Balance

* Indicates maximum value

The calculated values of the Hot Crack Sensitivity for given specifications, is shown in figure 9. Here results reveal that HCS for the typical SA240-304L is found least and is maximum for SA 312 (pipe) 316L. Hot cracking susceptibility increases markedly as the P and S contents exceeds 0.015%. Lowest value 0.053 is observed for P and S contents, which directly illustrates emphasis of P+ S effect on HCS. As with the increased value of P+S, HCS increases. The sulphur range should be 0.005 to 0.030% for acceptable penetration for both types of steels. Increasing the nitrogen content improves the resistance to hot cracking of those types of 304L and 316L steels, which contain no niobium as nitrogen containing steels have less hot cracking resistance.

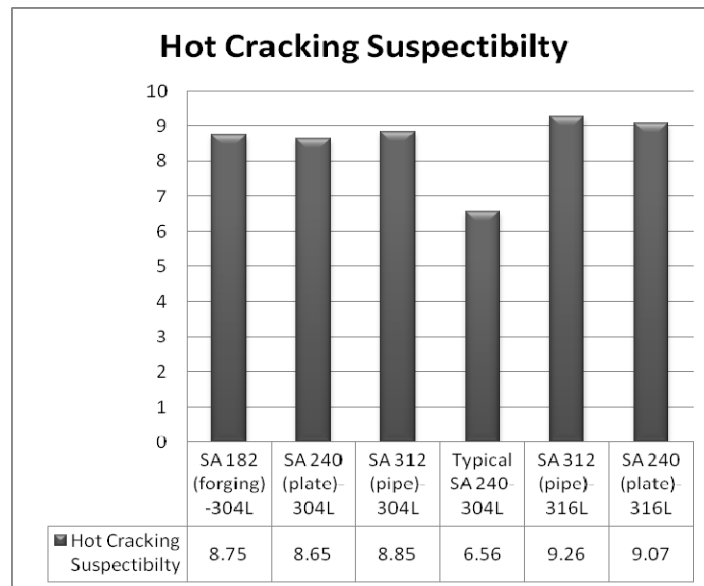


Fig. 9 Calculated HCS of various specifications of types 304L, and 316L using the extremes of the composition

9. Conclusion

A review of hot cracking in both steel welds shows that the problem is more prevalent in 316L steels. The propensity for hot cracking is determined primarily by detailed effects of impurity elements S, P and Si and alloying elements Cr, Ni, Mn and Mo for various compositions of 304L and 316L. The result presented here reveals HCS for the typical SA240- 304L is found least and maximum for SA 312 (pipe)-316L. The value of HCS was quite high for SA 312 (pipe)-316 L and SA 240(plate)- 316L steels in compare to type 304L steel. Hence filler metals should overalloyed with Cr, Mo, Si, Ni, C, N and Mn in most cases for controlling formation

of delta ferrite. Variations within the composition limits of these alloys can affect both the mechanical properties and welding characteristics of the alloy. Solidification mode was found major determinant of HCS, ensuring an FA or F mode ensures the best resistance to hot cracking. The final weld metal structure should contain a few percent delta ferrite, which is a sign of a sound weld.

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