The Relationship between Solidification Mode Ferrite Morphology and Fissuring in Austenitic Stainless Steel Weldments —A Review

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ABSTRACT

A small volume percentage of ferrite in the weldmetal need not necessarily ensure good weldability of austenitic stainless steels. The location and morphology of ferrite in the weldmental is the one which actually plays a major role in hot cracking resistance. This depends on solidification of the alloy and the nature of the solid state ferrite-to-austenite transformation. Added to this impurity elements can alter the relationship between ferrite content and cracking. Both Schaeffler and Delong diagrams predict the amount of retained ferrite in the weldmetal from the alloy chemistry. But both the solidification mode and the rate of cooling to room temperature affect the amount of ferrite retained. These two factors are sensitive to the welding process and welding parameters. Hence concluding about the weldability of austenitic stainless steels, only from compositional consideration is very much inaccurate and dangerous. This paper tries to bring out these aspects clearly by reviewing the essential research work done so far on these lines.

1. Introduction

Austenitic stainless steels have been used in various chemical, fertilizer and food industries. Lately, the demand for this as a structural material in the fields of nuclear power generation, coal liquefaction and cryogenics services has been accelerating with the energy problem. In this situation, there is a marked increase in the use of austenitic stainless steels. This has led to intense activity in research and development of stainless steels. To enhance the quality of these materials by the control of impurities and alloying elements, a lot of progress has been made in steel refining techniques. This progress should be accompanied by a similar advancement in the field of welding techniques. Hot cracking susceptibility of austenitic stainless steels depends upon metallurgical phenomena related to chemical composition and constitutes one of the major factors affecting weldability. Even though many studies have been made in this respect [1,2] they have not succeeded in thoroughly clarifying the mechanism of hot cracking. The potent effect of a percentage of ferrite in reducing hot cracking has been known for years and has resulted in the development of a number of diagrams, such as that published by Schaeffler and later revised by Delong et al which predicts weld ferrite content from chemical composition.

The relationship between ferrite content and weld cracking has led to the development of codes which require weld ferrite, above a specified minimum level in an attempt to eliminate cracking. However, in some cases designers would like to minimize or eliminate ferrite for reasons such as preventing transformation of ferrite to sigma phase, preventing preferential corrosion of ferrite or excluding magnetic phases. Therefore, conflicts often arise between the designer and welding engineer.

But recent works[3] have shown that requiring the existence of ferrite does not necessarily ensure good weldability. In particular, the location of the ferrite formed in the microstructure during the solidification process appears to be the dominant factor in determining weldability. In that case, acceptance or rejection of weldments based upon empirical ferrite calculations as found in various diagrams or by magnetic measurements such as those determined by Magne-Gage readings

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could prove unwise. For a given composition, ferrite may be present in various amounts and in different morphologies within the weld depending on the welding processes and parameters. Hence the prediction of ferrite content from strictly compositional considerations would appear to be inherently inaccurate.

Hence, to gain a full understanding of the role of ferrite in reducing hot cracking susceptibility of these alloys, one has to understand the solidification process of these weldments. This paper tries to bring these aspects into light.

2. Ferrite and Hot Cracking Resistance

The minimum weldmetal ferrite necessary in avoiding hot cracking can vary from less than 2% for 16-8-2 material for instance, to 4-10% for type 309[4-6]. Alloys which exhibit either a minimal amount of retained ferrite such as type 310, or an extreme amount such as type 312, are inherently susceptible to hot cracking. On the other hand, there is no correlation of amount of ferrite with the cracking susceptibility of the alloys which fall between these two extremes. A high ferrite content also seems to promote cracking so that there is an optimum range for the highest cracking resistance. This complex relationship shows that the ferrite content at room temperature can be the only factor involved in the cracking problem. Thus the amount of residual delta ferrite necessary to ensure an 'acceptable' level of hot cracking resistance is specific to a particular alloy composition, and unfortunately, may vary greatly within the nominal composition limits of that alloy.

3. Mode of Solidification

Within the composition limits for a specific type of stainless steel, the balance between ferrite and austenite stabilizers is the major factor controlling the solidification mode. When the ferrite stabilizers are dominant, delta ferrite is the first solid to crystalize from the liquid. Conversely, when austenite stabilizers predominate, austenite is the first solid to form from the liquid.

During cooling to room temperature, much of the primary delta ferrite becomes unstable and transforms to austenite. On the other hand, primary austenite remains stable during cooling to room temperature and only the interdendritic eutectic-ferrite, if any exists, partially or totally transforms to austenite.[7]

Four modes of solidification are possible in Fe-Cr-Ni stainless steels[8-13]:

(1) Solidification completely to delta ferrite

(2) Solidification completely to austenite

(3) Solidification as austenite with the formation of divorced eutectic ferrite at the dendrite interstices during the terminal transient stage of solidification.

(4) Solidification as delta ferrite until partitioning of austenizers to the remaining liquid causes this liquid to solidify as austenite.

The initial solidification product is dependent only on the nominal composition of the melt at the liquidus temperature. However, segregation of alloying elements during non-equilibrium solidification shifts the overall composition of the remaining liquid and alters the final solidification product. An increase in the concentration of austenizers in the remaining liquid or a decrease in the amount of ferritizers, favours solidification as austenite. Correspondingly, a local enrichment in ferrite forming elements ahead of the solid-liquid interface promotes solidification as delta ferrite.[14]

Since chromium and nickel are the principal alloying elements in austenitic stainless steels, the (chromium/nickel) ratio is the dominant factor in controlling whether solidification occurs as delta ferrite or austenite. In addition, manganese which is half as powerful as nickel in stabilizing austenite is added from 1% to 2% weight and silicon which is 1.5 times more powerful than chromium in promoting ferrite is normally present from 0.5 to 1.0 wt%. Thus the nominal amount of manganese and silicon has essentially equal and opposite effects and should have little combined effect on which phase is the first to solidify.

Carbon, which is present in amounts less than 0.1 wt percentage and nitrogen which may be picked up during the welding process are both powerful austenizers and tend to promote primary austenite solidifictaion. Trace elements like sulphur and phosphorus have little effect on the solidification mode, although segregation of these elements during freezing is the major cause of hot cracking.

When $Cr_{eq}/Ni_{eq} \leqslant 1.48$, the resulting microstructure is austenitic or austenitic ferritic. In this microstructure, the delta ferrite, if any, is interdendritic. A microstructure in which the delta ferrite is located mainly at the dendrite axes results when $1.48 \geqslant Cr_{eq}/Ni_{eq} \leqslant 1.95$. This is ferritic austenitic microstructure. Single phase ferritic solidification occurs when $Cr_{eq}/Ni_{eq} > 1.95$ resulting in a microstructure in which lath morphology dominates and the solidification substructure is normally invisible.

4. Solidification behaviour and Hot Cracking Susceptibility

Hot cracking is associated with the segregation of certain alloying elements to interdendritic and intergranular volumes during the solidification of welds. Hot cracking is believed to occur at or above the solidus temperature of the lowest melting phase present.[15,16] During final stages of solidification, narrow, solid bridges separating areas of low melting liquid are subject to the greatest proportion of the shrinkage-induced strains. An increase in the amount of low-melting phase or the inherent strain resulting from soldification shrinkage may cause fracture of these solid bridges, thus forming hot cracks. Investigations of fully austenitic weld metals have revealed a high incidence of hot cracking both in the fusion zone and in the base metal adjacent to the fusion zone. It has been widely established that tramp elements such as sulphur and phosphorus increase the hot cracking susceptibility of many stainless steels. It has also been known for years that a small volume percentage of ferrite in the room temperature microstructure of austenitic stainless steel weldment reduces the hot cracking sensitivity. Moisio et al have proposed that duplex mode of solidification increases the interphase interfaces between austenite and delta ferrite and minimizes the area of austenite-austenitie and delta-delta grain boundaries at solidification temperatures[17].

These grain boundaries must be rare when the weld solidifies in a ferritic-austenitic manner. They have also shown that the number of initial cracks does not depend on the solidification mode[18]. Hence nucleation of cracks is independent. High cracking resistance of welds solidified in the ferritic-austenitic mode is due to the difficulty of crack propagation, because phase boundaries between two different lattices are not wetted by liquid as easily as are grain boundaries between two similar lattices.

The importance of mode of solidification or the primary solidification from the liquid has been clearly shown by Masumoto et al[19]. In their experiments, fusion welds which solidified as primary austenite were susceptible to hot cracking, while those which formed primary delta ferrite were immune. Important of all, they found no correlation between cracking susceptibility and the ferrite content as estimated by the Schaeffler-Delong diagram. This is an extremely significant result against the concept of correlating the ferrite content at room temperature with the susceptibility to hot cracking.

During solidification as primary austenite, chromium, silicon, sulphur and phosphorus are rejected to the liquid and may exist in relatively high concentrations in the final solid which forms along the solidification boundaries. During solidification as primary delta ferrite, relatively less sulphur and phosphorus are rejected to the liquid. In addition, the last to solidify region is enriched in both manganese and silicon and thus these elements are likely to combine to form phosphides and sulphides.

The carbon was not found to segregate significantly either during delta or austenite solidification. In addition, the fact that the delta ferrite has a greater solubility for harmful elements such as sulphur and phosphorus than does austenite is postulated to reduce the volume of low melting constituents formed during solidification.

Alloys which contain a large proportion of chromium and/or nickel such as type 310 and 312, lie in closer proximity to the ternary eutectic point (49 Cr-43Ni-8Fe) and would be expected to form a larger proportion of eutectic constituents than alloys more remote from this composition, such as type 304, 304L, 308 and 316. A refined microstructure of the duplex weldments of type 304L, 316, 308 and 309 creates more boundary area over which low melting eutectic phases are distributed. As a result, it is unlikely that continuous films could exist. In addition, these alloys contain smaller proportions of chromium and nickel. Thus they would form a smaller amount of eutectic constituents along the solidification boundaries, which is beneficial from the hot cracking point of view.[20]

5. Ferrite Morphology

The strength and corrosion behaviour of austenitic stainless steel welds are influenced by ferrite in various ways.[21–24]. Apart from the hot cracking point of view, recently it has been found out that, the amount and morphology of ferrite influence the sensitization behaviour of duplex stainless steel[25]. For a particular carbon content, there exists a critical amount and distribution of delta ferrite-austenite boundary area, above which the alloy is immune to sensitization. Hence, the ferrite morphology plays a very important role in many of the weld behaviours.

It has been reported that the as-welded ferrite morphology is strongly dependent on the welding parameters.[26,27] In general, if the welding heat input is high, the solidification substructure tends to be coarsened and results in a more widely spaced ferrite network. Lower heat inputs are accompanied by faster cooling rates and promote the formation of finer substructures which provide the formation of finer ferrite network.

In addition, since the local cooling rate varies in a continuous fashion from the fusion line to the weld centre line, point to point variations in ferrite morphology are frequently observed within the weldmetal. Takalo et al[28] have reported that a direct correlation exists between the morphology and the relative amount of ferrite present in the as-welded microstructure. They propose that as the percentage of ferrite increases in the range from 5 to 15 volume percent the 'vermicular' morphology is gradually replaced by 'lathy' ferrite. However, solidification studies performed by a few others indicated that upon rapid cooling from solidification range, the as-welded vermicular structure was replaced by the same lath-like morphology. Hence, it is apparent that the cooling rate of the weldment exerts considerable influence upon the final ferrite morphology.

Referring to the schematic Fe-Cr-Ni pseudo-binary diagram (fig 1) Savage et al[29] has shown that it is possible to predict the behaviour of a wide range of Fe-Ni-Cr alloys upon cooling from solidification range. He has defined four specific compositional regions on the Fe-Cr-Ni pseudo-binary diagram each of which exhibits a characteristic ferrite morphology.

Region : 1. Alloys in this range, solidify as primary austenite and may form a limited amount of ferrite as a divorced eutectic along the intercellular boundaries.

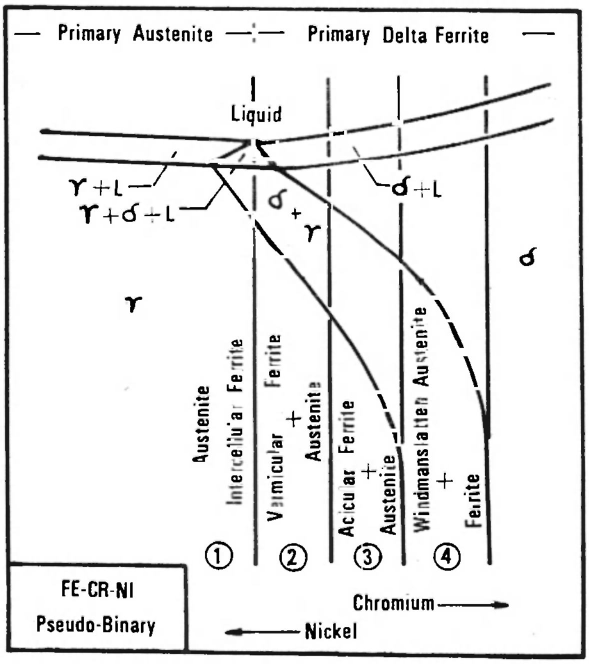

Fig. 1. Schematic pseudo-binary diagram of the Fe-Cr-Ni ternary system illustrating the effect of Composition on austenite ferrite morphology in austenitic stainless steel metal.

If the Cr/Ni ratio of this ferrite is sufficiently high to render it stable at room temperature, it usually exhibits a semi-continuous morphology.

Region : 2. Alloys in this range, solidify as primary delta ferrite dendrites whose cores are highly enriched in chromium and depleted in nickel. Upon cooling through the two phase (austenite plus delta) region, the ferrite of nominal composition formed during steady state solidification transforms to austenite by a composition-invarient, mass transformation. A portion of the ferrite at the dendrite cores is sufficiently enriched in chromium and depleted in nickel to remain stable at room temperature and is characterized by a vermicular morphology. As the (chromium/nickel) ratio increases within this region, the ferrite network becomes more continuous.

Region : 3. In alloys within this range, the primary delta ferrite is stable over a relatively large temperature range. But the cooling rates present suppress the diffusion-controlled transformation of the ferrite and the microstructure exhibits an acicular morphology at room temperature.

Region : 4. For alloys in this range, the pseudo-binary diagram predicts that ferrite and austenite should co-exist in a near-equilibrium mixture at room temperature. The composition of the austenite formed in these alloys differs from the nominal composition. Hence a massive transformation is impossible in these alloys. Consequently, a diffusion controlled transformation of ferrite to austenite must occur upon cooling through the two phase region. Thus, the as welded microstructure consists of ferrite and widmanstatten austenite which nucleates at the austenite grain boundaries and forms along specific habit planes in the ferrite. It should be noted that the position of the alloy composition within regions 1 to 4 does not preculde the formation of alternate microstructures. For example, an alloy whose composition lies near the boundary of region 2 and region 3 may exhibit both vermicular and acicular ferrite in the as-welded microstructure.

The rate at which the weldment cools through the two phase (austenite+delta) region can also have a significant effect on both the ferrite distribution and morphology. But weldmetal solidification rates encountered within the confines of common welding processes affect the amount of ferrite in the microstructure very little. Hence, near the fusion line where the cooling rate is the greatest, an acicular microstructure may be produced which is different morphologically from that of the microstructure in the interior of the fusion zone. A relationship between ferrite content and weldmetal microstructure exists[31]. But the thermal cycles to which some of the initial weld passes are subjected during a multipass weld are another factor that could contribute to variations within the weldmetal. Since ferrite in the weldmetal is not an equilibrium structure, thermal cycling could dissolve it, thus bringing a change in ferrite content of the weld from region to region. The distribution and morphology of delta ferrite in austenitic stainless steel weldments is dependent on the electrode and base metal compositions, welding parameters, the local weld cooling rate and the degree of dilution.

The location of delta ferrite in the microstructure can be used to determine whether the primary phase during solidification was ferrite or austenite. When the retained ferrite is located at the interstices of the cellular dendrites, the primary solid phase was austenite and the retained ferrite was the product of a divorced eutectic reaction during the terminal transient period of solidification. When the retained ferrite is located at the cores of the cellular dendrites, the primary solid phase was delta ferrite and the retained ferrite was formed during the initial transient period of the solidification process.[30]

6. Weld Composition and Hot Cracking Susceptibility

Weldmetal analysis can affect the cracking tendency both through the balance between 'austenizers' and 'ferritisers' and because some individual elements can have appreciable effects.

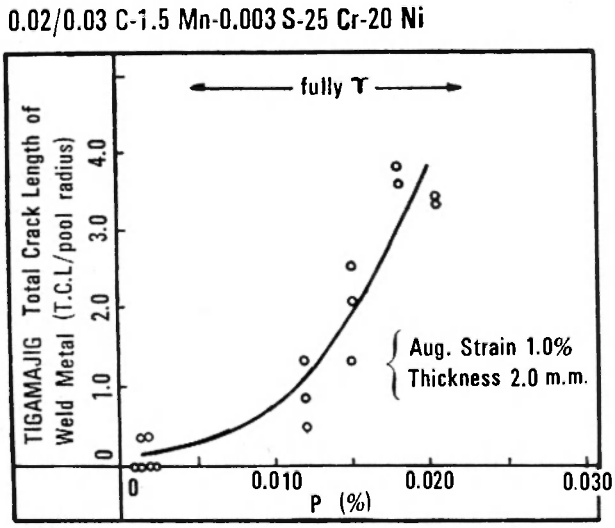

Fig. 2. Effect of P on hot cracking susceptibility in fully austenitic stainless steel (Ref. 32).

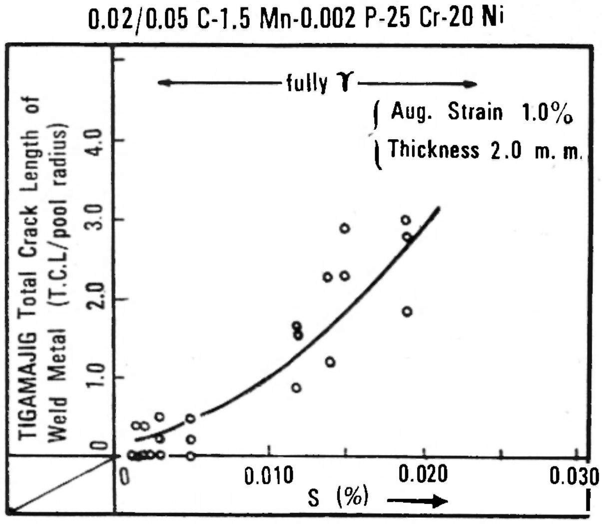

Fig. 3. Effect of sulphur on hot cracking susceptibility in fully austenitic weldmetal through varestraint test (Ref. 32).

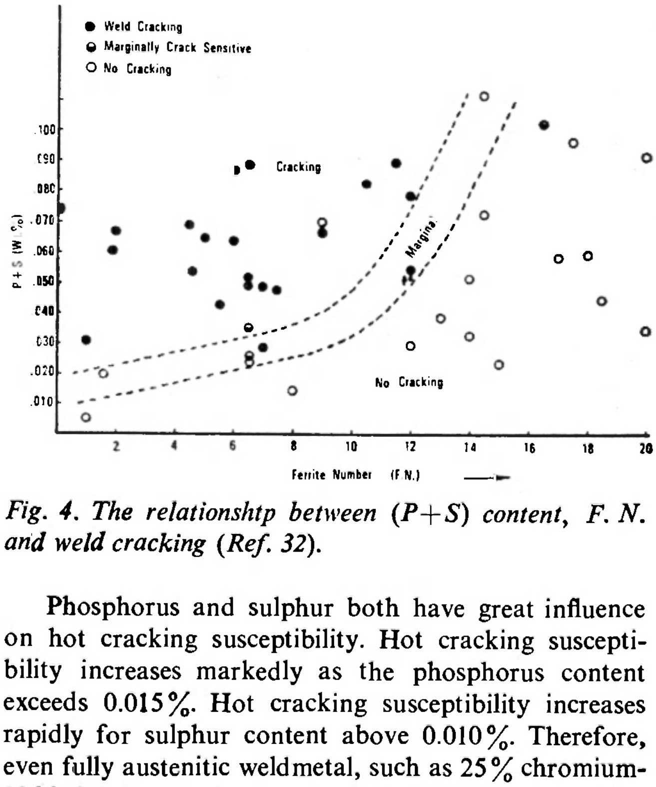

Fig. 4. The relationship between (P+S) content, F. N. and weld cracking (Ref. 32).

Phosphorus and sulphur both have great influence on hot cracking susceptibility. Hot cracking susceptibility increases markedly as the phosphorus content exceeds 0.015%. Hot cracking susceptibility increases rapidly for sulphur content above 0.010%. Therefore, even fully austenitic weldmetal, such as 25% chromium-20% nickel can be improved substantially in hot cracking resistance by lowering both phosphorus and sulphur contents to about 0.002%. When Cr_{eq}/Ni_{eq} is plotted against (phosphorus + sulphur) value, the graph indicates a definite value for the Cr_{eq}/Ni_{eq} 1.49 below which the welds are susceptible to cracking and above which they are not. This threshold value coincides almost exactly with the boundary between primary austenitic

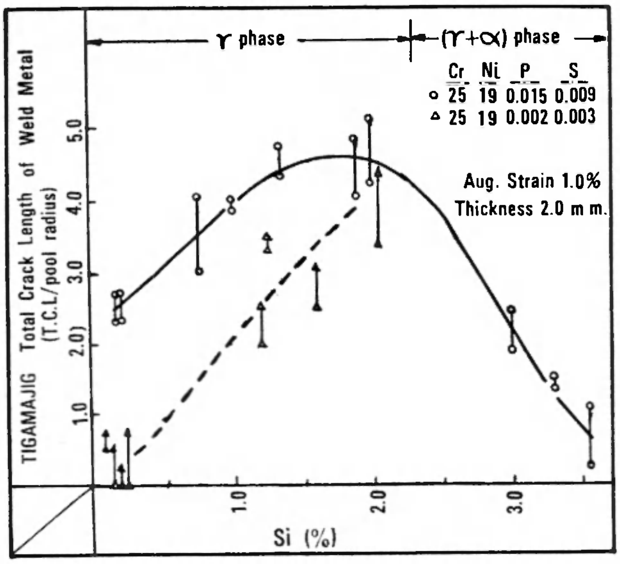

Fig. 5. Effect of Si a hot cracking susceptibility in austenitic weld metal (Ref. 32).

0.013 C-0.2 Si-1.4 Mn-0.003 P-0.011S-24 Cr-23 Ni

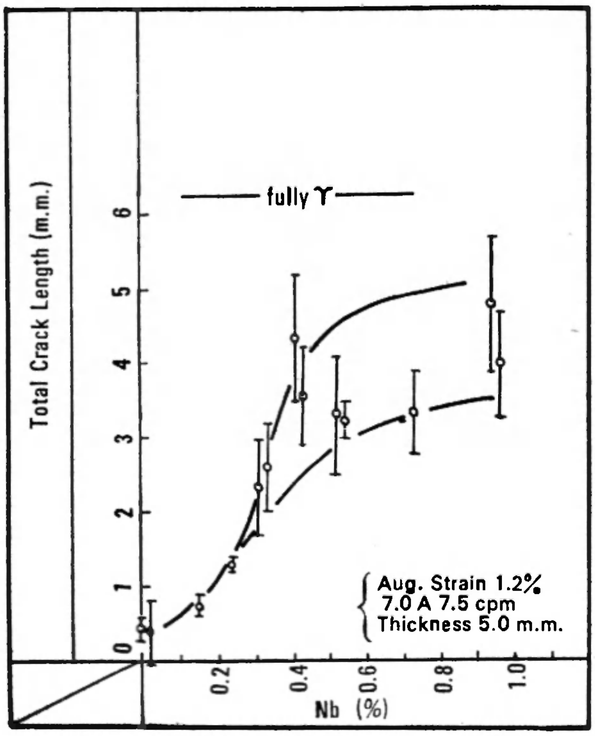

Fig. 6. Effect of Nb on hot cracking susceptibility of fully austenitic stainless steel weld metal (Ref. 32).

and ferritic solidification. Welds with a low impurity level (P+S≈0.01%) are nevertheless crack insensitive independent of their Cr_{eq}/Ni_{eq} ratio. The use of the sum (P+S) implies that the effects of the impurities are approximately equal (Fig. 4.)

0.20 Si-1.5 Mn-0.003 P-0.011 S-24 Cr.-23 Ni-0.04 N

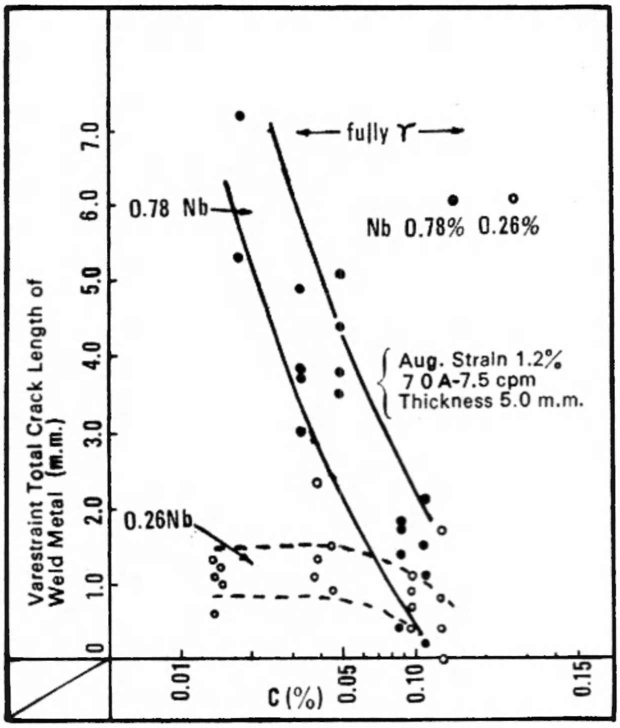

Fig. 7. Effect of C on hot cracking susceptibility of fully austenitic Nb-containing stainless steel (Ref. 32).

Silicon has an effect similar to that of phosphorus and sulphur on the hot cracking susceptibility of fully austenitic weld metal. Hot cracking suceptibility increases linearly with silicon content until approximately 1.5%, drops sharply beyond 2.0%, at which point delta ferrite begins to form and settles to a level in the vicinity of 3.5%[32] (Fig. 5).

As the niobium content exceeds 0.30% the hot cracking susceptibility of fully austenitic weldmetal increases rapidly (Fig. 6). The effect of carbon on the hot cracking susceptibility of 0.26% and 0.78% niobium containing steels are shown (Fig. 7). For 0.78% niobium steel, the hot cracking susceptibility decreases substantially with increase in carbon content. On the other hand, the hot cracking susceptibility of 0.26% niobium steel is low and relatively independent of carbon content below about 0.10% carbon.

The effect of nitrogen depends on the presence of niobium. Increasing the nitrogen content improved the hot cracking resistance of fully austenitic stainless steel

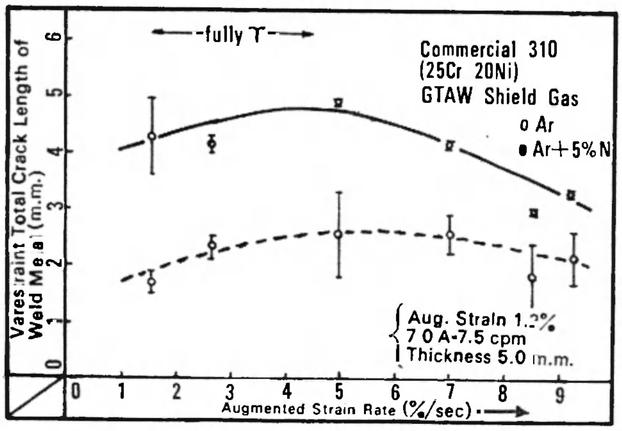

Fig. 8. Effect of N on hot cracking susceptibility in fully austenitic stainless steel weld metal without Nb. (Ref. 32)

0.02 C-0.20 Si-1.5 Mn-0.005 P-0.003 S-18.8 Cr-9.3/10 Ni-0.25 Nb

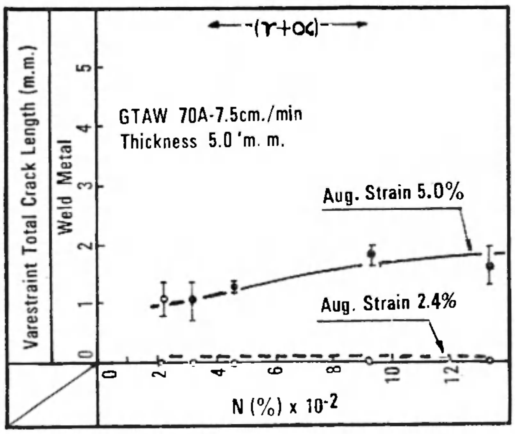

Fig. 9. Influence of N on hot cracking susceptibility of low C, Nb containing stainless steel with duplex (γ+α) weld metal. (Ref. 32)

which contains no niobium (Fig. 8). Nitrogen contents above 0.02% in niobium-containing steels decrease the hot cracking resistance significantly Figs. 9 & 10. Nitrogen increases hot cracking susceptibility slightly even in duplex welds containing niobium (Fig. 11). The effect of nitrogen is particularly important since increasing the arc length during welding increases the nitrogen content of the asdeposited weldmetal. Increasing manganese content from normal levels (generally 1-2 wt%) to 5-10% decreases hot cracking susceptibility as a result of its ability to combine with excess sulphur. However, there is some indication that increased manganese results in decreased corrosion resistance.

0.015 C-0.35 Si-1.65 Mn-17.6 Cr-11.3 Ni-0.015 P-0.30 Nb- 0.01 S

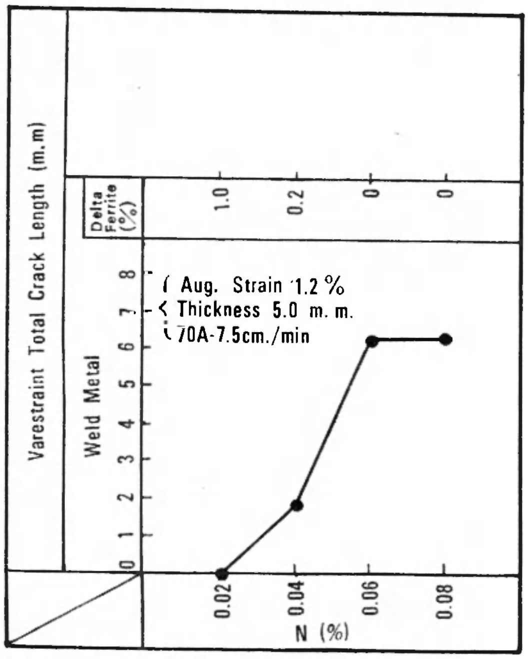

Fig. 10. Influence of N and δ ferrite on hot cracking susceptibility of low C, Nb containing stainless steel weld metal. (Ref. 32)

0.005 C-0.20 Si-1.5 Mn-0.004 P-0.003 S-18.6 Cr-9.7 Ni-0.025 N

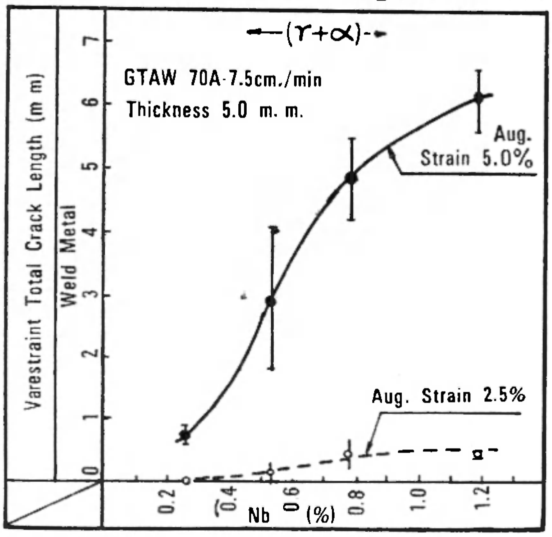

Fig. 11. Influence of Nb on hot cracking susceptibility of low C, N containing stainless steel with duplex (γ+α) weld metal. (Ref. 32)

7. Conclusions

(a) A very low as well as very high ferrite content in the weldmetal promotes hot cracking. There is an optimum range of ferrite level for the highest cracking resistance. This optimum range is specific to a particular alloy composition.

(b) The mechanism of hot-cracking has not been thoroughly clarified. In austenitic stainless steel welds, the mere existence of ferrite does not necessarily ensure good weldability. The location and morphology of the ferrite formed in the microstructure during solidification process is the one which determines weldability. Hence judging the weldability based upon empirical ferrite calculations from various diagrams or magnetic measurements may mislead.

(c) Four different modes of solidification are possible in Fe-Cr-Ni stainless steels. They are :

1. Fully austenitic
2. Austenitic-ferritic
3. Ferritic-austenitic and
4. Fully ferritic.

When Cr_{eq}/Ni_{eq} is less than 1.48, the resulting microstructure is austenitic or austenitic ferritic. In this the delta ferrite is interdendritic. When Cr_{eq}/Ni_{eq} is between 1.48 and 1.95, the resulting microstructure is ferritic austenitic. The delta ferrite is located at the dendrite axes. When the ratio is more than 1.95, single ferritic phase solidification occurs.

(d) Welds which solidify as primary austenite are susceptible to hot cracking, while those which form primary delta ferrite are immune. There seems to exist no correlation between ferrite content and hot cracking susceptibility.

(e) A direct correlation exists between the morphology and the relative amount of ferrite in the as-welded microstructure. The cooling rate of the weldments exerts considerable influence upon the final ferrite morphology. Depending on the mode of solidification, there exists four distinct ferrite morphologies in the as welded austenitic stainless steel weldments. They are:

(1) discontinuous vermicular ferrite
(2) continuous vermicular ferrite
(3) acicular morphology and
(4) widmanstatten austenite.

From the ferrite morphology the mode of solidification can be inferred.

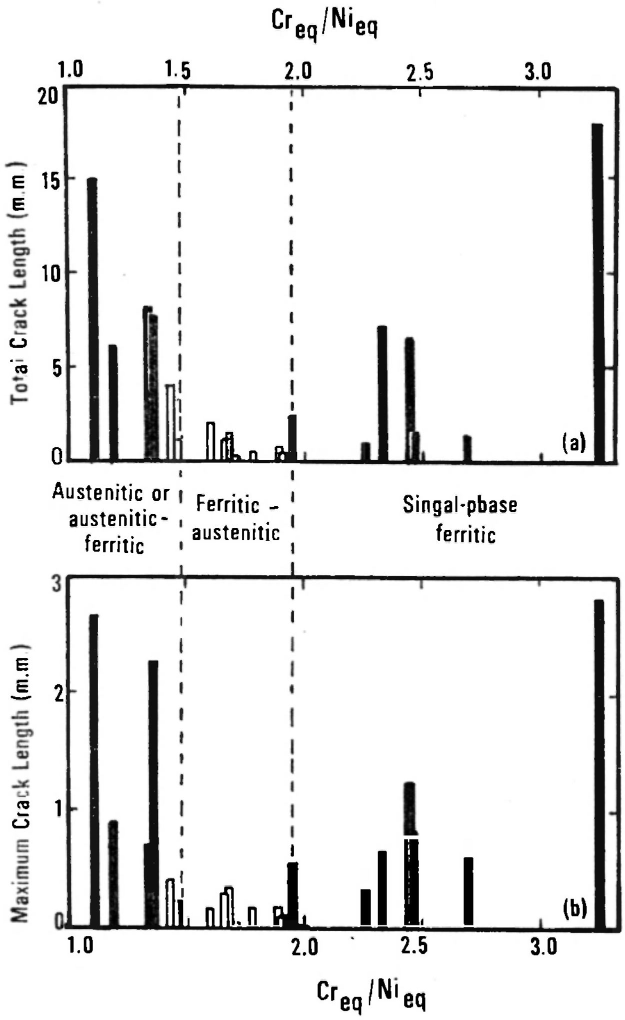

Fig. 12. Results of transvarestraint tests in increasing order of the ratio Cr_{eq} Ni_{eq} in order of solidification mode. (Ref. 17)

(f) The amount of retained ferrite in weldmetal and its morphology are sensitive to the welding process and its parameters. Hence, predicting ferrite content from strictly compositional considerations is not accurate.

(g) Silicon and niobium increase the hot cracking susceptibility. Carbon improves the hot cracking resistence only in steels containing niobium. Nitrogen improves the hot cracking resistance of fully austenitic stainless steel containing no niobium. In the case of niobium containing steels, nitrogen decreases the hot cracking resistance.

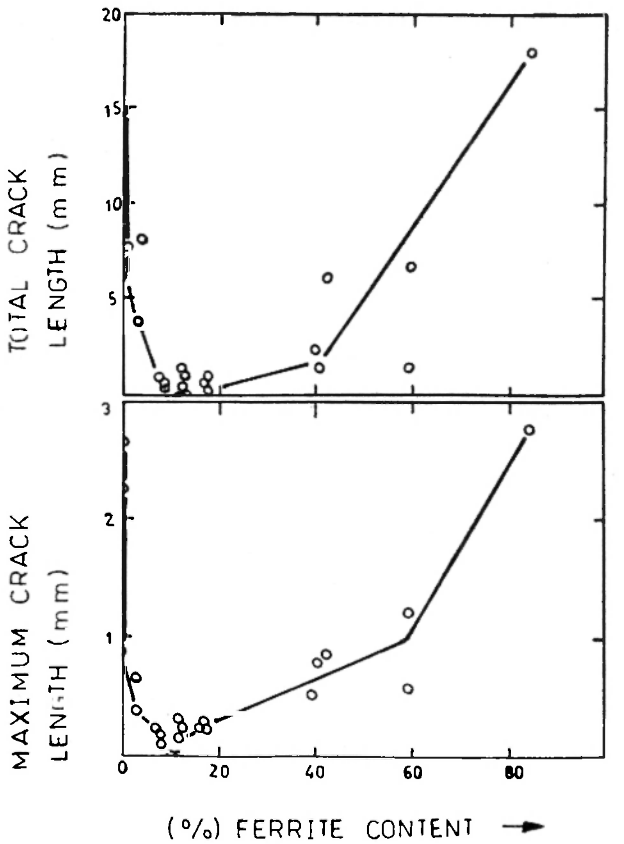

Fig. 13. Results of transvarestraint tests as a function of measured ferrite content. (Ref. 17)

(h) While both sulphur and phosphorus are very detrimental to hot cracking, the ferrite level that prevents hot cracking is very much dependent upon these impurity levels.

References

1. Borland, J. C. ; Suggested explanation of hot cracking in mild and low alloy steel weld. Welding Research Abroad, 1962, VIII (2), 73-87.

2. Hemsworth, B ; Boniszewski, T ; and Eaton, N.F ; Classification and definition of high temperature welding cracks in alloys. Metal construction and Brit, Weld. J., 1969 ; 2, 5s-16s.

3. Cieslak, M. J. and Savage, W. F. ; Weldability and solidification phenomena of cast iron alloy heat resistant HK-40 and corrosion resistant 19Cr-9Ni-2.5Mo steels. Alloy Casting Institute Project, A-69, 1979, Aug.

4. Lundin, C. D. ; Weld. J. ; 1975, 54(8), 241s-246s.

5. Brooks, J. A. and Lambert, F. J. ; Weld. J., 1978, 57(5), 139s-143s.

6. Brouwer, G ; Philips Weld. Rep., 1978 (3), 16-19.

7. Cieslak, M. J. ; Ritter, A. M. and Savage, W. F. ; Solidification cracking and analytical electron microscopy of austenitic stainless steel weldmetals, Weld. J., 1982, 61(1), 1s-8s.

8. Cieslak, M. J. and Savage, W. F. Weld. J. 1980, 59(5), 136s-146s.

9. Arata, Y ; Matsuda, F and Katayama, S ; Trans. of JWRI, 1976, 5(2), 35-51.

10. Lippold, J. C. and Savage, W. F., Weld. J., 1980, 59 (12), 362s-374s.

11. Arata, Y ; Matsuda, F and Saruwatari, S ; Trans. of JWRI, 1974, 3(1), 79-88.

12. Lyman, C. E., Weld. J., 1979, 58(7), 189s-194s.

13. David, S. A. ; Goodwin, G. M. ; and Braski, D. N., Weld. J. 1979, 58 (11), 330s-336s.

14. Lippold, J. C. and Savage, W. F., Solidification of austenitic stainless steel weldments : Part 1-A Proposed Mechanism, Weld. J, 1979, 59 (12), 362s-374s.

15. Pellini, W. S., Strain theory of hot tearing, The Foundry, 1952, 80, 125-133.

16. Borland, J. C. ; Generalized theory of super-solidus cracking in welds and castings, British Weld. J., 1960, 7, 508-512.

17. Kujanpaa, V. P., Suutala, N. J., Takalo, T. K. and Moisio, T. J. I., Solidification-cracking estimation of the susceptibility of austenitic and austenitic-ferritic stainless steel welds, Metal Construction, 1980, 12(b), 282-285.

18. Kujanpaa, V. P. ; Suutala, N. J, Takalo, T. K. and Moisio, T, Welding Research International, 1979, 9(2), 55-76.

19. Masumoto, I ; Tamaki, K ; and Kutsuna, M ; Hot cracking of austenitic steel weldmetal, Trans. JWS, 1972, 41(11), 1306-1314.

20. Lippold, J. C. ; and Savage, W. F. ; solidification of austenitic stainless steel weldments ; Part III-The effect of solidification behaviour on hot cracking susceptibility, 1982, 61(12), 388s-396s.

21. Delong, W. T. ; Ferrite in austenitic stainless steel weldmetal, Welding J., 1974, 53(7), 273s-286s.

22. Hauser, D ; and Vanecho, J. A. Properties of steel weldments for elevated temperature pressure containment applications, MPC-9, Smith, G. V. ed., Amer. Soc. of Mech. Engg., New York, 1978, 17-46.

23. Edmonds, D. P. ; Vandergriff, D. M. and Gray, R. J., Ibid, 47-61.

24. Castro, R and De Cadenet, J. J., Welding Metallurgy of stainless and heat resisting steels, Cambridge University Press, London, 1974.

25. Devine, T. M., Influence of Ferrite Morphology and Carbon content on the sensitization of duplex stainless steel, Abstract, 109th AIME Annu. meeting, Las Vegas, Nevada, Feb. 1980.

26. Goodwin, G. M. ; Cole, N. C. ; and Slaughter, G. M., A Study of ferrite morphology in austenitic stainless steel weldments, welding J., 1972, 51(9), 425s-429s.

27. Asakura, S ; Wachi, H ; and Watanabe, K ; The effect of welding conditions on the crack sensitivity in austenitic stainless steel weldments, Trans. JWS, 1972, 3(2), 34-44.

28. Takalo, T ; Suutala, N ; Moisio, T ; Influence of ferrite content on its morphology in some austenitic weldmetals, Met. Trans., 1979, 10A(4), 512-514.

29. Lippold, J. C. and Savage, W. F., solidification of austenitic stainless steel weldments, Part II, The effect of alloy composition on ferrite morphology, Welding J., 1980, 59*(2)*, 48s-58s.

30. Cieslak, M. J. and Savage, W. F., Ferrite morphology in high molybdenum stainless steel, welding J., 1981, 60*(7)*, 131s-134s.

31. Cieslak, M. J. and Savage, W. F., Weldability and solidification phenomena of cast high alloy heat resistant IIK-40 and corrosion resistant 19Cr- 9Ni-2.5 Mo steels, Alloy Casting Institute project, 1979, Aug., A-69.

32. Ogawa, T ; and Tsunetomi, E ; Hot cracking susceptibility of austenitic stainless steels, Welding J., 1982, 61*(3)*, 82s-93s.

33. David, S. A. ; Ferrite morphology and variations in ferrite content in austenitic stainless steel welds, Welding, J., 1981, 60(4), 63s-71s.

34. Brooks, J. A. and Lambert, F. J., The effects of phosphorus, sulphur and ferrite content on weld cracking of type 309 stainless steel, welding J., 1978, 57(5), 139s-144s.

35. Discussion on "Ferrite morphology" in high molyodenum stainless steels", Welding J., 1982, 61(5), 164s-165s.