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HAZ Cracking in Thick Sections of Austenitic Stainless Steels—Part II

Metallurgical reactions involving cracking mechanisms, composition, and other effects are reviewed in detail

BY R. D. THOMAS, JR.

ABSTRACT. This is the second of a two part document, the first reviewing the occurrence of cracking, the laboratory methods of study, and the effect of fabrication variables. This part deals with the metallurgical reactions which help to explain the phenomenon and draws conclusions which are pertinent to both parts of the document.

Microstructural transformations are temperature dependent, many of which are time dependent. By the use of electron microscopy, the principal forms and size of carbides, which result in liquation cracking during grain boundary migration, can be observed. Certain carbides and other compounds of residual or trace elements dissolve at temperatures approaching the solidus and form a low melting eutectic with the matrix austenite in the grain boundaries; this is unable to resist the strains occurring during solidification of the weld metal.

Electron microscopy permits the identification of the constituents that form during reheating, suggesting that reheat cracking or cracks formed during service at elevated temperature are the result of carbide precipitation. Those which form within the grains cause a strengthening; those which form in the grain boundary denude the nearby region, causing a weakening. Decohesion at the grain boundary is observed in such cases.

Despite a considerable number of studies, there remain areas of doubt as to the principal causes of cracking and the means to control it. It has been suggested that columbium, which imparts creep resistance to austenitic stainless steels, might be substantially reduced with little or no sacrifice to the creep properties and with greatly improved resistance to heat-affected zone cracking. Research in this area and in others relating to the composition of the steels is needed, especially that relating to the control of residual and trace elements.

Finding reliable solutions to this problem will benefit the users and producers of austenitic stainless steels. Another result will be to shed light on heat-affected zone cracks encountered in nickel-base alloys where the problem frequently occurs in substantially thinner sections.

See pages 24 to 32 in this issue of the Welding Journal *for Part I of this paper.*

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Introduction

The first part of this two-part document dealt with the nature of cracks, which are found in thick section austenitic stainless steel weld heat-affected zones. The cracks are found sometimes during fabrication, and often after months or years of service at elevated temperatures.

Considerable research has been published during the past thirty years. Much of this involves observations of the metallurgical reactions which occur at elevated temperatures. The reactions are seen as short time effects as during the welding cycle or during postweld heat treatment. They are also observed as long term effects of exposure in service, as in electric generating facilities. It is the metallurgical reactions which are primarily dealt with in this part of the document.

Theories of Cracking Mechanisms

The observation that heat-affected zone cracks are almost always intergranular suggests that compositional segregation of some sort may be responsible. The series of papers presented at an ASM seminar in 1977 (Ref. 50) provides a broad background for the theories being offered by various researchers.

Most writers (Refs. 6, 7, 8, 51, 52, 53) treat the question of the mechanism of cracking as a liquation phenomenon. Both Tamura (Refs. 51, 52) and Savage (Ref. 53) consider that compositional segregation in the as-received base material is due to lack of complete homogenization and shows up as a banded structure.

These localized variations in solute content account for the variability of different grain boundaries to show evidence of liquation. Moreover, solute segregation accumulates in grain boundaries during rapid heating to temperatures above 2372°F (1300°C) when grain boundary migration occurs, "sweeping" the solute-rich alloy to new grain boundaries. Tamura found that, in Type 310 stainless steel, a heating rate of 252°F (140°C) per second caused chromium (Cr) and nickel (Ni) enrichment at the grain boundaries roughly double that in the grain matrix. But when the alloy had been completely homogenized to remove any solute-rich banding, solute segregation at the grain boundaries was eliminated.

Grain boundary migration is impeded by the presence of precipitates. In Type 347 stainless steel, large columbium carbide (CbC) particles on further rapid heating form a low melting CbC-austenite eutectic which appears to wet the grain boundaries. This suggests that improved results could be obtained if large CbC particles could be removed by a solution heat treatment and subsequently given a stabilizing heat treatment to precipitate the carbide in as fine a form as possible. Titanium carbide (TiC) particles in Type 321 stainless steel, which also form a eutectic but at a higher temperature than CbC eutectics, dissolve more readily than CbC; hence, they are less likely to be present as large particles. This, according to Tamura, accounts for the lesser sensitivity of Type 321 stainless steel to cracking than that of Type 347 stainless steel.

The presence of other forms of precipitates may be either harmful or beneficial, depending on their action on the liquating grain boundaries. Residual or tramp elements, which migrate to the solute-rich grain boundary, are frequently the cause of heat-affected zone cracks. The presence or absence of other metallic phases, especially ferrite, is frequently suggested as influential in liquation cracking.

More complex explanations are offered when considering the "ductility dip" form of cracks. In general, these have not been as widely researched to provide a clear statement of the cracking mechanism. In Type 347 stainless steel, the most common explanation of ductility dip cracking is the strengthening of the grains due to fine intragranular precipitates; this allows no possibility for the strains to be accommodated within the grains and forcing grain boundary decohesion. This explanation is generally accepted for the cracking which occurs in postweld heat treatments (see Part I).

Guttmann and McLean (Ref. 54) suggest that there may be strong interaction between impurity and major alloying elements. Phosphorus (P), for example, which is now recognized to interact with chromium in low alloy ferritic steels to cause temper embrittlement, may similarly cause an embrittling phenomenon in austenitic stainless steels. Yet, these authors state, "It must be confessed that no evidence of metallic segregation [of phosphorus] has been given yet for austenitic steels, but very few experiments have been performed for the moment, and the above examples have amply shown that very careful and extensive measurements and sputter profilings are necessary when the segregation ratio is smaller than, or of the order of, two." They go on to state, "... and grain boundary decohesion must be accepted as endangered by segregation even at high temperatures.... Moreover, in slow fracture at elevated temperature, fresh segregation can occur to cover the new surfaces as they are exposed, and the total work for the separation of the interface is lowered.... As far as the austenitic steels are concerned, the knowledge provided by the theory of coupled segregation described in this paper makes unwise the continued neglect of this possibility."

Effect of Composition

A thorough understanding of the role of the individual elements in the stainless steel on their influence on heat-affected zone cracking is clearly needed. Certain elements, such as columbium (Cb), have been thoroughly researched and are reasonably well understood. Other elements appear to affect the susceptibility adversely in certain situations and favorably in others.

Combined effects of elements, such as the role of manganese (Mn) in reducing the adverse effect of sulfur, have received some but comparatively little attention. Trace elements are invariably referred to in the published literature, but systematic studies of their effects are noticeably lacking. Phase constituents, especially ferrite, whose effect in the control of weld metal fissuring is well recognized, have been the subject of several investigations, but deserve additional attention.

The following paragraphs attempt to

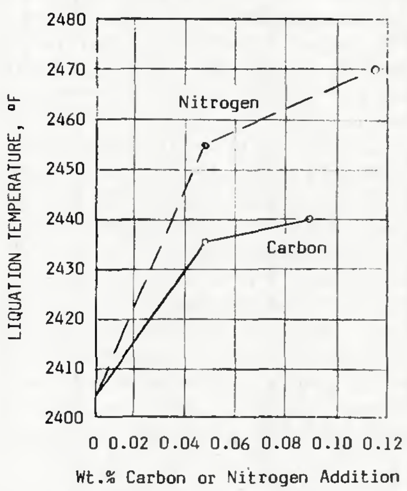

Fig. 1—Liquation temperature of Type 347 stainless steel welds as a function of carbon or nitrogen addition to an alloy containing 0.055% C and 0.77% Cb (Ref. 55)

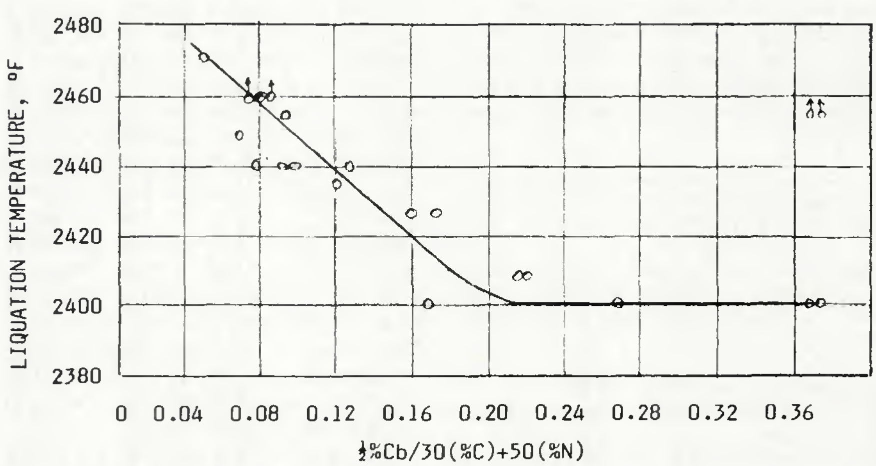

Fig. 2—Liquation temperature of Type 347 stainless steel as a function of the ratio of columbium content to the carbon and nitrogen contents (Ref. 55)

summarize the areas of agreement, as well as the lack of agreement, on the role of elements and phase constituents.

Carbon and Nitrogen

These two elements are treated by a great number of researchers. Cullen and Freeman (Ref. 55) examined the liquation of the CbC eutectic in Type 347 stainless steel, arriving at the relationships shown in Fig. 1. When carbon (C) is reduced leaving free columbium to form the low melting eutectic Fe_2Cb, the liquation temperature is lowered. To prevent this, carbon must exceed the stoichiometric ratio of CbC (see discussion of columbium below). They proceed to discuss the effect of nitrogen (N) on the liquation due to the Cb(C,N) eutectic which results in the graph shown in Fig. 2. This is presented as an empirically derived observation, which, the authors state, "needs additional heats with nitrogen for verification."

Morishige (Ref. 59) in a statistical analysis of 40 heats of austenitic stainless steels derives an equation to measure the liquation cracking susceptibility as measured by hot ductility tests:

$$\Delta H = -700C + 17Cr - 37Ni - 117Nb + 29Mo + 188 \quad (1)$$

As shown in Fig. 3, when ΔH is greater than 100, "the material exhibits an excellent hot ductility." This would seem to be contrary to the finding of Cullen and Freeman (Ref. 55), where columbium-bearing alloys show an increase in liquation temperature with a reduction in carbon. The apparent discrepancy is accounted for in that Morishige's Type 347 stainless steel alloys involve many heats which were partially ferritic, whereas the Cullen findings are based primarily on fully austenitic alloys. The Morishige's

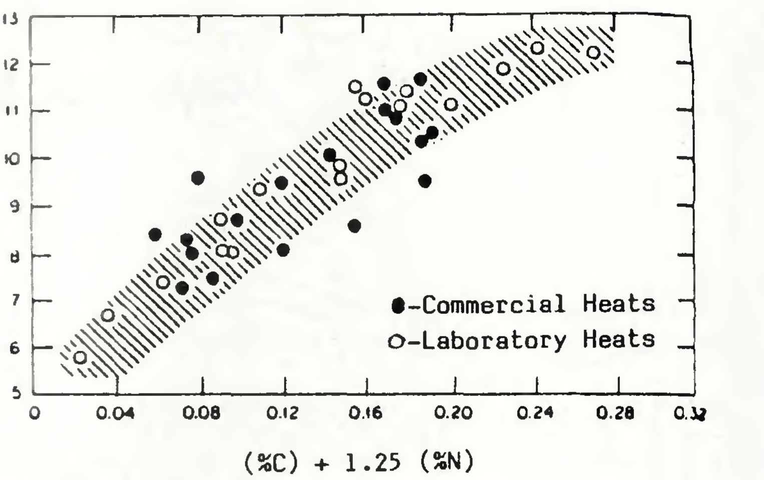

Fig. 4—Influence of carbon and nitrogen on the 100,000-hour rupture strength of Type 304 stainless steel at 1202°F (650°C)

equation, therefore, is influenced by the role of the elements in producing ferrite (see the section below on the role of ferrite on liquation).

In a series of Varestraint tests, Arata (Ref. 56) noted a difference in the augmented strain required to produce heat-affected zone cracks between Types 316 and 316L stainless steel heats, the latter being somewhat less susceptible than the former. From this one experiment, it would appear that 0.026% carbon is preferable to 0.07% carbon. The nitrogen contents of the two heats were not reported. This observation is confirmed in a review by Robinson (Ref. 7), citing Honeycombe (Ref. 57) and Sadowski (Ref. 58).

Using the Spot Varestraint (Tigamajig) test, Ogawa (Ref. 60) observed both fusion zone and heat-affected zone cracks, the latter being relatively minor. Nevertheless, the authors noted a small increase in heat-affected zone cracks with an increase in nitrogen from 0.02 to 0.06% in an essentially fully austenitic Type 347 stainless steel sample containing 0.015% C and 0.30% Cb. In a series of partially ferritic Type 347 stainless steel compositions with 0.02% C, 0.25% Cb, and nitrogen varying from 0.02 to 0.13%, the crack length in the heat-affected zone rose slightly to a maximum at 0.09%, after which it appeared to decline. In all of these tests, however, the total heat-affected zone crack length was less than 1 mm (0.04 in.), raising some doubt as to the significance of the results.

While carbon and nitrogen may appear to have an adverse effect on the heat-affected zone cracking of non-columbium-bearing alloys, it must be recognized that they have a favorable effect on the stress-to-rupture properties. Swindeman (Ref. 61) recently reviewed published and other work on the influence of residual and trace elements on the creep properties of Types 304 and 316 stainless steels.

Figure 4 shows the combined effect of carbon and nitrogen on Type 304 stainless steel, based on the work of Goodall (Ref. 62). A computer analysis of commercial heats of Type 316 stainless steel reported by Lai (Ref. 63) suggested that subject to "further restrictions from welding and fabrication considerations," the limits for carbon should be restricted to from 0.03 to 0.06%, and, for nitrogen, from 0.04 to 0.07% for improved creep properties.

The direction of some of the efforts to produce steels with high resistance to heat-affected zone cracking is to minimize carbon and nitrogen (the latter being probably the less harmful) and to find other elements to provide creep strength. For example, Moorhead and Sikka (Ref. 64) have produced laboratory heats of 0.05% C and 0.03% N with only

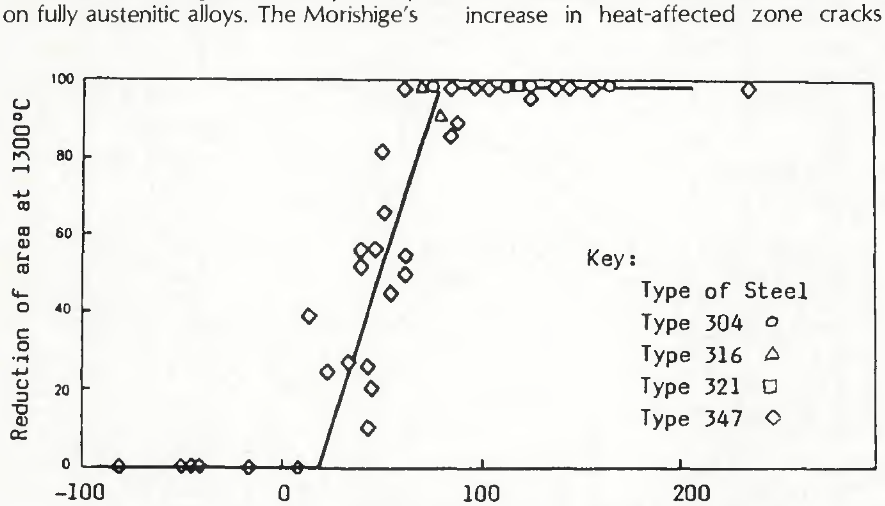

Fig. 3—Relationship between hot ductility parameter ΔH and reduction of area, measured at 2372°F (1300°C) after heating to 2462°F (1350°C) (Ref. 59)

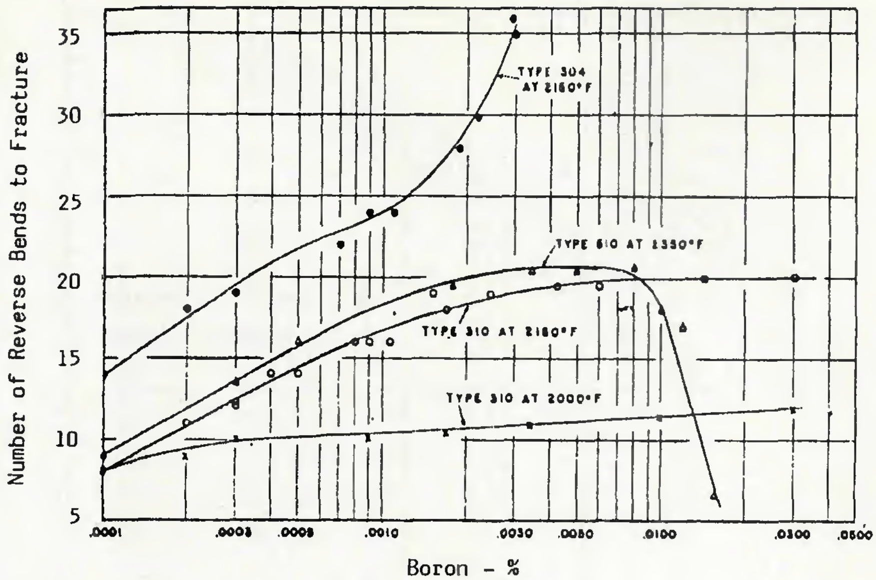

Fig. 5 — Effect of boron on the hot ductility of Types 304 and 310 stainless steel (Ref. 65)

0.10% Cb to give creep properties comparable to Type 347 stainless steel and crack susceptibility comparable to Type 304 stainless steel. This seems also to be the direction taken by the Japanese as indicated by the report of Ogawa discussed above.

Boron

Closely related to carbon and nitrogen in their metallurgical reactions is boron (B). It is not surprising that many researchers study boron along with carbon and nitrogen.

In 1957, boron was recognized as a desirable addition to stainless steel to overcome "hot shortness." Loveless and Bloom (Ref. 65) found additions of 0.002 or 0.003% improved remarkably the number of reverse bends to fracture at temperatures ranging from 2000 to 2300°F (1093 to 1260°C) in Types 304 and 310 stainless steel — Fig. 5.

Goldschmidt (Ref. 66) studied the solubility of boron in 18Cr, 15Ni austenitic stainless steel — Fig. 6. The eutectic form of the boride in the high temperature region is M_2B. Tamura (Ref. 51) in a discussion of liquation cracking indicated that $M_{23}(C,B)_6$, M_3B_2, and Ni_4B_3, among others, can liquate and induce heat-affected zone cracks. Brooks (Ref. 39) claimed that the reduction of boron from 0.006 to 0.001% greatly reduced hot cracking tendency in the heat-affected zone of alloy A286. In two later papers Brooks and colleagues (Ref. 40, 67) point out the beneficial effect of boron on stress-rupture properties, but its amount must be limited to avoid heat-affected zone cracking.

Donati *et al.* (Ref. 68) observed liquation in the heat-affected zone when boron exceeded 0.001% in Type 321 stainless steels. In a subsequent study (Ref. 69), they found that, although boron up to 0.0035% could be kept in solution by rapid quenching, this was not considered industrially feasible. That document also reported on tests of Type 316 stainless steel with boron varying from 0.0005 to 0.0112%. Under 0.0035%, the normal annealing practice will maintain the boron in solution, or as an intragranular $FeMo_2B_2$, which will not cause heat-affected zone cracking. Subsequently, work from the same laboratory by Boudot (Ref. 70) showed that the boron limit for Type 316 stainless steel may be as high as 0.0045% for electric furnace steels, but must be held to below 0.0025% for vacuum melted steels. This is because some boron is tied up as an inoffensive compound in electric furnace melting, but not in vacuum melting.

The motivation for the study of boron and its effect on heat-affected zone cracking arises from the improvement in creep properties by the addition of a small amount of boron. Williams (Ref. 71) found improved creep-rupture life and ductility in Type 316 stainless steels with 0.003 and 0.005% boron as compared with those having lower boron contents. They attributed these superior results to the formation of $M_{23}(C,B)_6$ particles with a higher boron to carbon ratio, which inhibits wedge cracks and creep cavities.

Lai (Ref. 63) observed from his computer study that the optimum boron for Type 316 stainless steel should be 0.002 to 0.005%. In a study of Type 308 stainless steel weld metal, Edmonds (Ref. 72) noted that boron appeared to increase rupture life and ductility; Types 308 and 316 stainless steel weld metals are proposed to contain controlled residual elements (CRE), the optimum boron level being 0.006%, along with 0.04% phosphorus, plus 0.05% titanium for shielded metal arc welds, 0.2% titanium for sub-

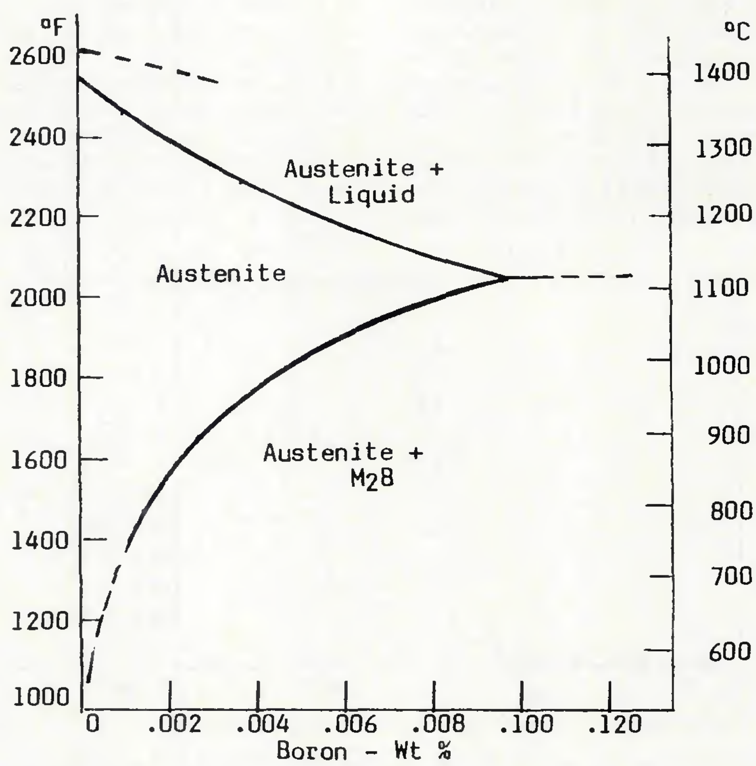

Fig. 6 — Solubility of boron in 18Cr-15Ni stainless steels (Ref. 66)

merged arc welds, and 0.5% titanium for gas tungsten arc welds.

Columbium (Niobium) and Titanium

These two elements, which are responsible for improved creep properties in stainless steels, are frequently cited as the principal causes of heat-affected zone cracking. Columbium and titanium, essential elements in Types 347 and 321 stainless steels, respectively, both form similar carbides preferentially to chromium carbides. This was their original purpose — to resist chromium depletion and consequent intergranular corrosion. As such, the standards for these grades are based upon the ability of the added elements to tie up enough carbon to make the steels suitable for resisting aqueous corrosion attack. But that is often not the need in applications requiring creep resistance. Consequently, the question arises as to whether the standards (*e.g.*, Cb 10 times C minimum and Ti 6 times C minimum) are appropriate for applications where creep is the primary consideration.

A study of the minimum columbium content for creep and corrosion resistance was conducted by Sikka (Ref. 73). This showed that 0.05% Cb in Type 304 stainless steel gave creep resistance comparable to Type 347 stainless steel (Fig. 7) and improved corrosion resistance by a factor of 5 over Type 304 stainless steel, although not quite so immune to intergranular attack as Type 347 stainless steel — Fig. 8. Moorhead and colleagues (Ref. 64, 74) claim to have demonstrated that with columbium reduced to 0.10% or less, the heat-affected zone is no longer susceptible to cracks.

A classic study by Cullen and Freeman (Ref. 55) has already been discussed above under the subheading "Carbon and Nitrogen." Liquation temperatures reach a minimum at increasing columbium levels, depending on the carbon content — Fig. 2. Excess columbium above this minimum is found in the eutectic as Fe_2Cb, which dissolves rapidly during heating and accounts for the reduced liquation temperature. Few other investigators have made note of this Fe_2Cb phase, except by reference to this classic work published in 1963.

As reported above under the subheading "Boron," Edmonds (Ref. 72) found that the small addition of 0.05% titanium to Type 308 stainless steel weld metals greatly improves the creep rupture life and ductility.

Auger electron spectroscopy was employed by Ogawa (Ref. 60) to show that columbium segregated to the grain boundaries where cracks were observed in the heat-affected zone. This confirms many earlier studies of standard Type

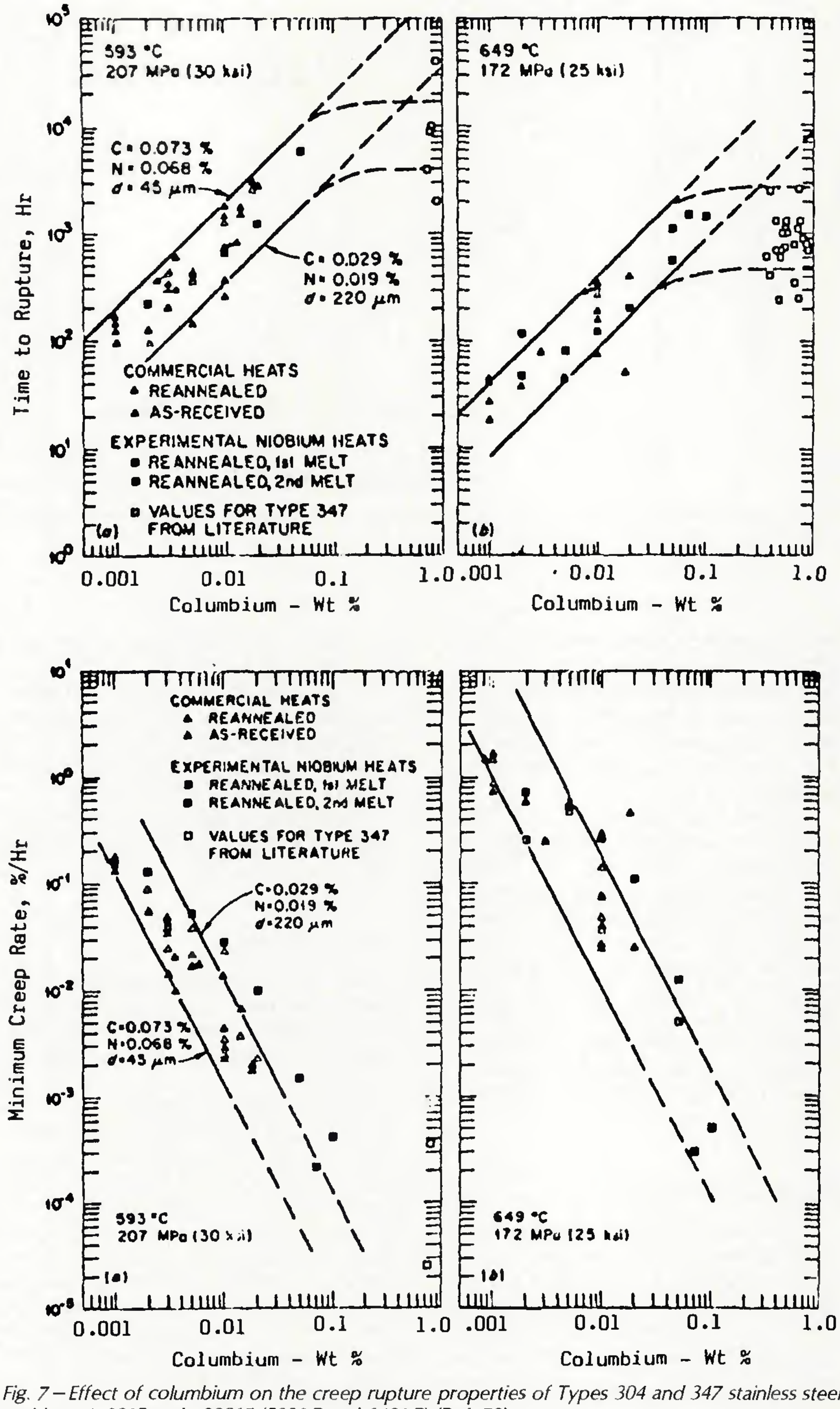

Fig. 7 — Effect of columbium on the creep rupture properties of Types 304 and 347 stainless steel welds at 1100°F and 1200°F (593°C and 649°C) (Ref. 73)

347 stainless steel. They found that this segregation and cracking occurred with columbium as low as 0.25% in a steel containing 21% Cr and 22% Ni, and 0.24% in a steel containing 24% Cr and 23% Ni.

Morishige's (Ref. 59) equation (1), cited above, assigns a negative coefficient of 117 to columbium, indicative of its harmful effect on hot ductility.

Significant segregation of titanium at grain boundaries was found by Lippold (Ref. 75) in alloy 800. He indicates that, as grain boundary migration occurs in the heat-affected zone, the carbides dissolve leaving a Ti-rich solute which offers preferential sites for crack formation under the cooling stresses. If the carbides in the alloy are present as relatively large particles, less solution occurs than if the parti-

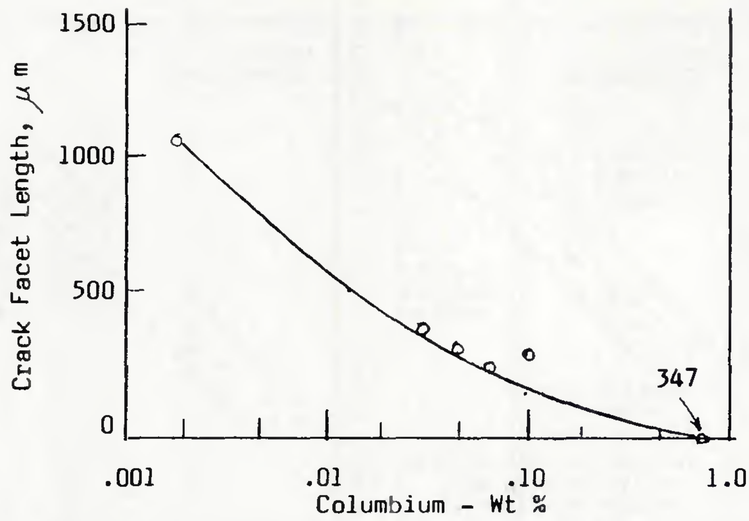

Fig. 8 — Effect of columbium on the intergranular corrosion of Type 304 stainless steels compared with one heat of Type 347 stainless steel (Ref. 73)

cles are small, causing relatively less heat-affected zone cracking. Thus heat-to-heat variations previously noted by other investigators are thought to be affected more by the prior annealing and mechanical treatments than by compositional variations.

Offering essentially the same cracking mechanism as Lippold, Brooks (Ref. 67) reported that increasing the titanium from 2.2 to 2.6% in A286 alloy decreased the zero strength temperature in the hot ductility test by almost 100°F (56°C). In an earlier study on A286, Brooks (Ref. 39) identifies the Laves phase, Fe_2Ti, which forms on the decomposition of the eutectic TiC liquated grain boundaries, resulting in grain boundary weakness considerably below the solidus temperature.

Manganese and Silicon

Their presence in all stainless steels subjects these elements to consideration as to their influence on the heat-affected zone. Silicon (Si) has long been considered a contributor to fusion zone hot cracking and microfissuring in fully austenitic weld metals. Manganese (Mn), on the other hand, has been looked upon as helpful in minimizing microfissures in weld metal. It is not surprising, therefore, to see these same effects being found in the heat-affected zone, although the number of reports giving significance to these elements are few.

As noted above, Brooks (Ref. 39) attributes cracking in A286 alloy to the Laves phase formation and shows that silicon enrichment of that phase lowers the zero strength temperature by 50°F (28°C). Using microprobe and Auger analyses, Lippold also observes an increase in silicon, along with titanium, in the grain boundaries of alloy 800.

In fully austenitic stainless steel, Ogawa (Ref. 60) points out the need to keep silicon low to reduce hot cracking in the fusion zone, but makes no claim that it is harmful in the heat-affected zone.

The beneficial effect of manganese is generally attributed to its tying up residual sulfur.

Sulfur and Phosphorus

Most writers consider both these elements detrimental, although there are few systematic studies of their adverse effect on heat-affected zone cracking. Persson (Ref. 76) is an exception, in that his study involved four heats of a 15% Cr, 15% Ni, 1.2% Mo,Ti,B austenitic stainless steel with sulfur(S) varying from 0.002 to 0.013%. In one of the high sulfur heats, 0.011%, cerium (Ce) was added to show its effect in tying up the sulfur. Grain boundary precipitates of titanium carbosulphide (sometimes referred to as tau-phase) increase in volume with increased sulfur — Fig. 9. These precipitates reduce the hot ductility when tested at temperatures from 2012 to 2372°F (1100 to 1300°C).

In a review of the effects of trace elements, Mayer (Ref. 77) cites Pruger (Ref. 78) as claiming that phosphorus can be tolerated up to 0.045% without adverse effect on hot workability. DeLong (Ref. 17) reported that phosphorus in the range of 0.02 to 0.03% was needed for creep properties.

In the microprobe analyses of the grain boundaries of alloy 800, Lippold (Ref. 75) found a considerable increase in phosphorus relative to the matrix alloy, but no increase in sulfur. This suggests that phosphorus may be a contributor to heat-affected zone cracking tendency in this alloy even though the matrix content of the element is less than 0.01%.

There are numerous references to the hot cracking of the fusion zone, indicating that both sulfur and phosphorus promote hot cracking in the absence of ferrite during solidification. Sulfur and, to a lesser extent, phosphorus partitions during solidification to the ferrite phase, removing them from segregating to the austenitic dendrite or grain boundaries where strains imposed by the solidification of subsequent weld passes could cause microfissures.

Nakamura and Abe (Ref. 79) discussed the means to produce steel with low sulfur and phosphorus contents, using what they term the MSR process. This is essentially an electroslag melting process using a CaF_2 slag bath in which calcium is dissolved. Sulfur and phosphorus are reduced to levels below 0.005%. Yet, in one of the diagrams presented by Ogawa (Ref. 60), the heat-affected zone cracking of the MSR melted steel is only marginally better than the commercial heat and another low phosphorus and sulfur heat, all of the Type 310 stainless steel analysis.

Copper

Little has been written on the effect of minor amounts of copper (Cu) on heat-affected zone cracking in austenitic stainless steels. There are standard austenitic stainless steel compositions in which copper is intentionally added as an alloying element in the range of 3 to 4%. Heat-affected zone cracking in such steels has not been reported. It has been speculated by some in private conversations that small amounts of copper (*e.g.*, under 1%) may be worse than larger amounts. No experimental evidence has been developed on this point.

Nevertheless, it is significant that Hull (Ref. 80) in his series of solidification cracking tests found the presence of up to 2% copper in fully austenitic stainless alloys to be detrimental, suggesting that copper does segregate during solidification. Hence, in the heat-affected zone thermal cycle, it may be presumed that copper segregation may be detrimental.

Copper is clearly detrimental to the heat-affected zone when present as a surface contaminant. Matthews (Ref. 81) and others (Refs. 82–84) have shown that such contamination has caused heat-affected zone cracks in many types of steels and other alloys by a mechanism termed "liquid metal embrittlement." As little as 0.003 mil (0.76 μm) of copper on the surface, such as might be caused by the abrasion of a copper hold-down fixture, or other copper tooling, could result in liquid copper penetrating the grain boundaries of the heat-affected zone and causing cracks during the cooling strains. Yet 0.96% copper in solution in the cobalt-base alloy L605 showed no sign of heat-affected zone cracks.

It might be noted also that the investigations at Rensselaer Polytechnic Institute during the 1940's and 1950's (Refs. 24, 85, 86) are reported to have been conducted on Type 347 stainless steels having 0.20 to 0.30% copper. There is no indication in these references that the copper was a harmful ingredient.

Other Low Melting Tramp Elements

Lead, bismuth, tin, silver, arsenic and antimony are tramp elements that have been found injurious when hot working stainless steels, especially hot rolling and hot forging of ingots. It is natural to suspect such elements when considering heat-affected zone cracking. Like copper, however, evidence is lacking which precisely identifies the role of these elements in grain boundary segregation during the welding thermal cycles.

Mayer and Clark (Ref. 77), in a review paper, cite references showing that:

1. 0.005% lead causes edge cracking when stainless steel ingots are forged or rolled.
2. Bismuth is ten times more potent than lead in this regard.
3. Silver, arsenic and antimony also have adverse effects.
4. Above 0.045% tin is detrimental to the surface quality of stainless steel sheet and strip.

Cullen and Freeman (Ref. 55), in their assessment of the causes of low hot ductility in Type 347 stainless steel, write, "There is a possibility that trace elements which are known to cause hot cracking can be present in the alloys. This investigation did not identify any such cases. No known cases of this have been reported. This investigation has been limited to cases of low hot ductility induced by normal compositional effects. It is quite certain, however, that with the information resulting from this investigation to define the role of normal compositional variations it would be possible to show that trace amounts of lead, tin, silver, etc., could be contributing factors to poor hot ductility."

Rare Earth Metals and Others

The literature fails to reveal any systematic study of heat-affected zone cracking from heats treated and untreated with rare earth metals (REM). The Morishige equation (1) cited above has been modified to take REM into account, assigning a positive coefficient of 3000 to this variable (Ref. 87):

$$\Delta H = -700C + 17Cr - 37Ni - 117Nb + 29Mo + 3000REM + 170 \quad (2)$$

In this modification, hot ductility is said to be favorable when ΔH is equal or greater than 60. The relation is considered applicable to Type 347 stainless steels where REM is under 0.02%. According to the table of compositions used for this statistical study, only two of the 36 heats were given rare earth treatments.

In the discussion of the paper by Donati *et al.* (Ref. 68), Mr. Tricot, of the French steel producer, Ugine, pointed out that hafnium is considered efficacious in combatting heat-affected zone cracks in boron-containing superalloys.

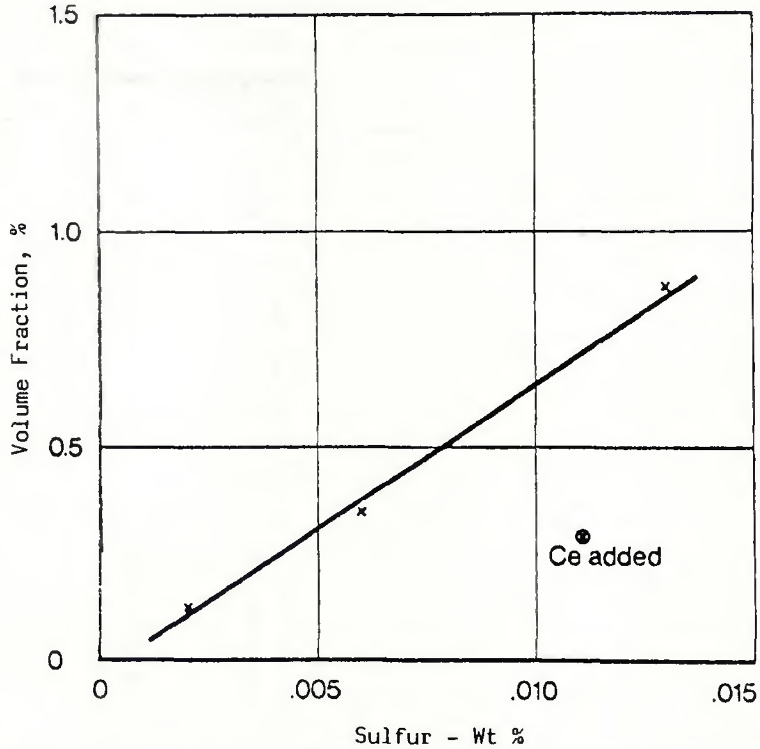

Fig. 9 — Volume fraction of Tau-phase appearing after heat treatment at 2507°F (1375°C) as a function of the sulphur content of a 15Cr, 15Ni steel (Ref. 76)

Oxygen and Oxide Inclusions

Little or no direct evidence identifies oxygen or its compounds with cracks in the heat-affected zone. Most low melting point oxides will have been removed in the steel melting practice, and high melting point oxides would be detrimental only if they nucleate low melting constituents during grain migration in the welding thermal cycle. This has not been suggested in any of the investigations, except by Tamura (Ref. 51), who includes "oxide type inclusions, possibly some low melting point silicates and spinels" among the phases which can liquate and induce grain boundary liquation.

Delta Ferrite

The single most powerful deterrent to solidification cracking is the presence of delta ferrite. Much has been written on this subject, because the principal problem of cracking in stainless steel welds has occurred in the fusion zone rather than in the heat-affected zone. But it is natural that investigators should include the presence or absence of ferrite in austenitic stainless steels when attempting to interpret compositional effects which determine the performance of a steel with respect to heat-affected zone cracking.

Wrought steel producers have long been aware of the benefit to hot workability by keeping the alloy "balanced" — that is, free from ferrite. This is illustrated in a paper by Bloom (Ref. 88), showing the relation between hot twist ductility and ferrite level of various stainless steels at temperatures from 2100 to 2400°F (1149 to 1316°C).

Most commonly used austenitic stainless steels in wrought form, especially those produced in thick sections, are of a composition which borders on the austenite-ferrite boundary. Although ferrite may be present in many of the heats in the ingot form, the subsequent hot work-

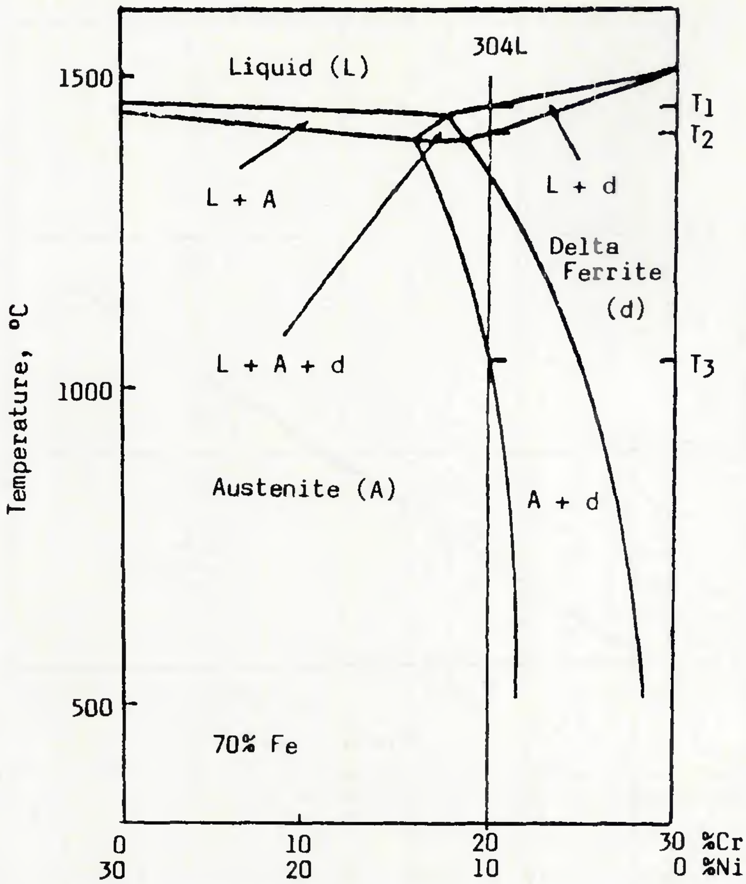

Fig. 10—Pseudo-binary slice of the Fe-Cr-Ni ternary phase diagram taken at 70% Fe

ing redissolves the ferrite, giving a uniform ferrite-free austenite.

The tendency of a given composition, which is fully austenitic in the wrought condition, to form ferrite during solidification under welding conditions can be evaluated by using an autogenous tungsten arc weld pass on the surface of the steel. The ferrite, if present, can be identified by metallographic techniques, or more simply by magnetic measurements, such as a Magnegage.* Most of the investigators who consider ferrite as a factor in evaluating the heat-affected zone cracking tendency of a specific heat use one of the constitution diagrams for weld metals or mathematical models which relate composition of the principal elements to ferrite tendency.

It would be well to note at this point that, although welds are commonly made in thick sections of cast austenitic stainless steels, there has been relatively little heat-affected zone cracking reported. This has been attributed to the fact that most commonly produced cast austenitic stainless steels almost invariably contain ferrite, even in the solution annealed condition in which the castings are usually supplied. Lundin and Spond (Ref. 10) have shown that microfissures in the heat-affected zones of previously deposited weld passes are reduced or eliminated when ferrite is present.

Figure 10 shows the phase diagram frequently presented to explain the formation of delta ferrite during solidification. But it can be equally valuable for showing the phase changes on the heating cycle in the hot ductility test or in the heat-affected zone. Since solid phase partitioning of elements is a relatively slow process, it may not take place during rapid heating. At any rate, the attempts to relate ferrite tendency to heat-affected zone cracking have met with varying success.

In the early work on hot ductility testing, Soldan (Ref. 5) observed ferrite in a commercial heat that performed well and found no ferrite in one that performed poorly. But in the same study, two other heats were found to remain fully austenitic, one of which performed well, the other poorly. There is a confusion in the report, namely that the metallographic evidence of ferrite in the two samples which showed a difference in ferrite were from samples subjected to a peak temperature of 2450°F (1343°C) and the two samples showing no ferrite were subjected to a peak temperature of 2400°F (1316°C). The authors conclude, nevertheless, that "it would seem that ferrite cannot be the cause of the difference in behavior during welding."

The liquation temperatures of a series of Type 347 stainless steels were determined by Cullen and Freeman (Ref. 55) covering a wide range of compositions and ferrite contents. Their study showed, "Increasing amount of ferrite in the structure . . . . was associated with first a slight decrease in liquation temperature and then a marked increase in comparison to wholly austenitic alloys with the same C and Cb contents. . . . Liquation associated with the depression of liquation temperature by small amounts of ferrite was confined to the interface of the ferrite grains with the austenite. When the ferrite content was about eight percent or above, the liquation temperatures were high, if liquation occurred at all below the bulk melting temperature. . . . In high columbium heats with small amounts of ferrite, Fe_2Cb formed at the interface of ferrite and austenite in the as-cast ingots."

On the last point cited above, Cullen and Freeman went on to observe that "columbium is 13 to 16 times more soluble in ferrite at the temperatures considered in their investigation." Thus on cooling, a columbium-enriched austenite should result as the ferrite dissolves, and the products should contain the precipitate form, Fe_2Cb. But when the ferrite content was sufficiently high that the columbium did not saturate it, then there was no rejection of the columbium to the interface with austenite, thus accounting for the high liquation temperature when ferrite exceeded about eight percent.

Morishige (Ref. 59), in his statistical study of 39 commercial heats, determined the ferrite content of a tungsten arc weld pass on each. No correlation could be found between the ferrite forming tendency of the heat and the hot ductility test—Fig. 11.

On far fewer heats, Donati (Ref. 89) concluded that the heat-affected zone may benefit from a ferrite forming tendency in alloys which generally have a wide solidification range, *e.g.,* boron-containing Type 316 stainless steel + Ti. His observations, however, are based on only five heats of different types, making it difficult to draw more than preliminary observations.

Sadowski (Ref. 58) found that steels which form a small amount of ferrite in the heat-affected zone are less susceptible to heat-affected zone cracks. In his review, Robinson (Ref. 7) quotes this reference on the influence of ferrite tendency.

Tamura (Ref. 52) observed the course of liquation during grain boundary migra-

**"Magnegage" is a trade name of the Magnegage Sales & Service Company.*

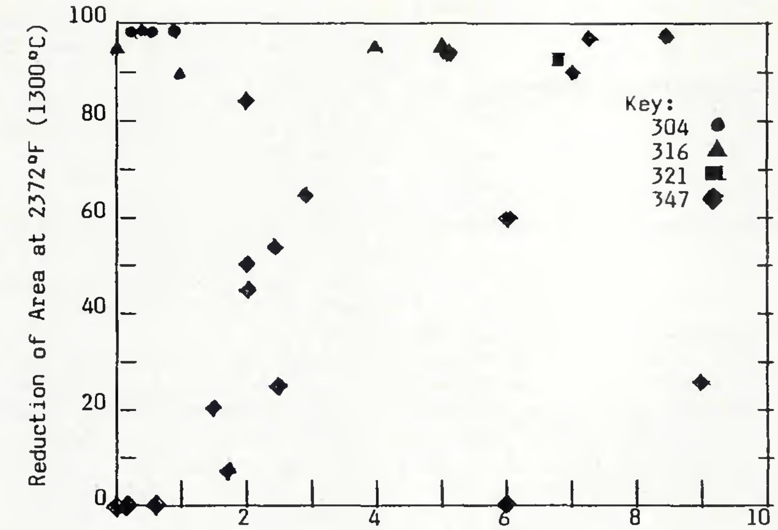

Fig. 11—Hot ductility as a function of the delta ferrite tendency (Ref. 59)

tion by means of a high temperature microscope. When bands of delta ferrite were present in a Type 347 stainless steel heat-affected zone, the initial liquation at carbide particles was joined by liquation of the delta ferrite, extending the grain boundary liquation and increasing the risk of cracking. He notes that this observation is at variance with the beneficial effect of ferrite in the fusion zone. He concludes, "... it seems that narrow ferrites extending in the rolling direction which have existed since room temperature exert a serious adverse influence. According to this discussion, it is necessary in the production process of these stabilized type steels to give such a heat treatment as to make these ferrites perfectly disappear."

Other Metallurgical Effects

As investigators seek to analyze their observations, it becomes apparent that composition is not the only criterion by which one can judge the weldability of stainless steels with respect to heat-affected zone cracking. A few of the other factors which are referred to in the literature are strain, diffusion, grain size, and back filling.

Strain-Induced Precipitation

Irvine (Ref. 21) does not deal specifically with heat-affected zones in welded construction, but rather with ductility of Type 347 stainless steel in stress-rupture tests. Strain-induced precipitation of CbC within the grains accounts for the low ductility of stress-rupture specimens at temperatures between 1292 and 1652°F (700 and 900°C). Plastic strain creates dislocations on which intragranular precipitates form, creating a "denuded zone" at grain boundaries which become increasingly weaker. As the test temperature increases, the size of the CbC precipitates increases, over-aging occurs, and the grain becomes weaker and more accommodating to accept the strains without placing too great a load on the grain boundaries. The strain is thus more uniformly distributed.

Younger (Refs. 19, 20) points out that, with a lower strength weld metal, less strain is placed on the heat-affected zone, causing fewer precipitates to form. This confirms the observation of earlier U.S. researchers who empirically found that the use of the comparatively weaker 16-8-2 filler metals was an advantage when welding crack sensitive heavy section austenitic stainless steels.

Residual stresses in thick-section welds are known to reach yield strength levels. Such strains accelerate the strain-age precipitation, and account for the reported cracking which has occurred in postweld heat treatments and in service.

Columbium and titanium form carbides of the CbC and TiC types, which are particularly harmful in strain-age hardening. In the absence of these elements, the usual form of carbide is $M_{23}C_6$ which forms both intra- and intergranularly, with much less hardening to the grains and simultaneously strengthening the grain boundary. Younger and Baker (Ref. 20) note that the specific volume ratio of CbC to the austenitic matrix is 1.91 compared with 1.76 for TiC and 1.15 for $M_{23}C_6$; they claim this accounts for the relatively high degree of crack sensitivity of the Types 347 and 321 stainless steels as compared with Types 304 and 316. They also suggest that these specific volume ratios can be altered by alloying the austenite with elements such as cobalt, manganese, and molybdenum, thus mitigating the problem.

The ductility dip is especially marked in

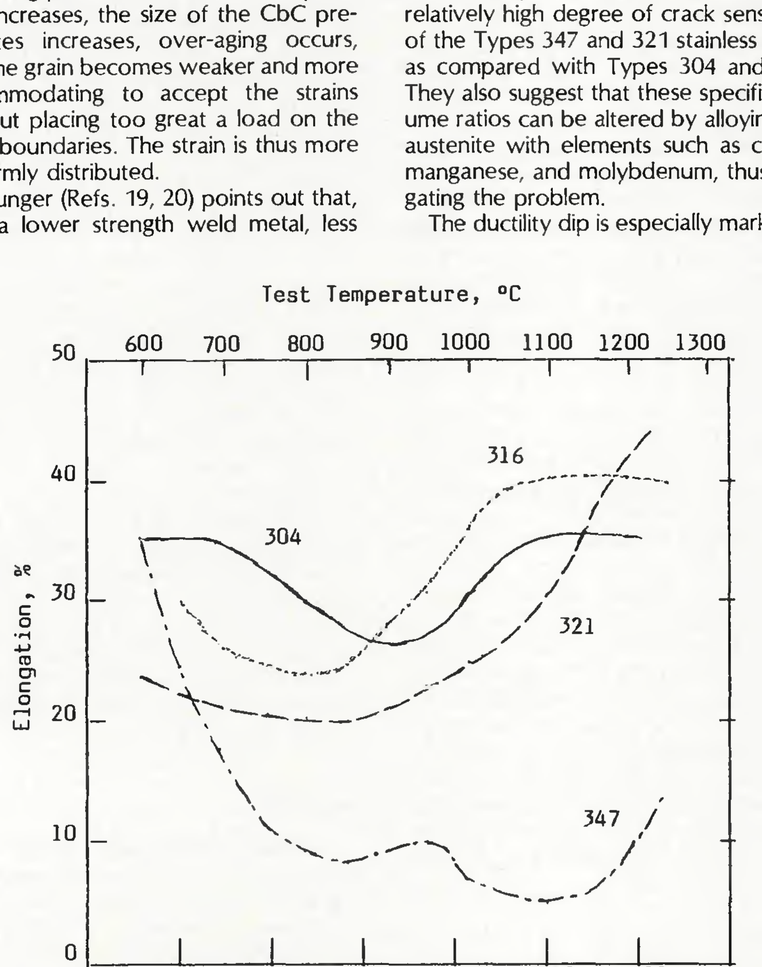

Fig. 12—Tensile ductility of various steels in short-time tests at elevated temperatures—Type 347 stainless steel air cooled from 2462°F (1300°); all others air cooled from 2507°F (1375°C) (Ref. 29)

Table 1—Diffusivity of Ti, Cr, and C in Austenite (Ref. 75)

Element	Alloy System	Diffusivity, $cm^2/s^{(a)}$ 1000°C	1350°C
Ti	Fe-Ti	3.8×10^{-11}	7.7×10^{-9}
Cr	Fe-17Cr-12Ni	8.4×10^{-12}	1.2×10^{-9}
C	Fe-15Cr-35Ni	8.5×10^{-7}	4.0×10^{-6}

(a) 1000°C = 1832°F; 1350°C = 2462°F.

Types 347 and 321 stainless steels. It has, however, been shown to occur also in Types 304 and 316 stainless steels (Ref. 29)—Fig. 12. This suggests that the latter steels, although much less prone to heat-affected zone cracking in thick sections, may be subject to it—especially Type 316 stainless steel operating in the creep range. It also explains, for these steels, the relative freedom from cracking problems during postweld heat treatments, and the merit of removing plastic strains by a postweld heat treatment.

Diffusivity

The rate at which reactions occur causing grain boundary weakness is discussed by several investigators to explain their observations.

In studies of liquation cracking, Tamura (Ref. 51) found that at heating rates less than 31°F (17°C) per second, diffusion of the solute-rich regions into grain matrix occurs during grain boundary migration sufficient to eliminate liquation at the grain boundary. Thus, depending on the elements which cause liquation, the rate of diffusion and the limiting rate of heating will differ.

Irvine (Ref. 21) indicated that the controlling effect in the loss of ductility caused by CbC precipitation is the rate of diffusion of columbium in austenite. This rate is considerably higher in regions of high dislocation density resulting from plastic strain.

In Lippold's study of alloy 800 (Ref. 75), the high diffusivity of carbon relative to chromium and titanium (Table 1) causes it to migrate rapidly to locations where titanium or chromium accumulates; this, he notes, is in the grain boundary of a weld heat-affected zone. By heating a specimen to 1832°F (1000°C), carbides are formed with titanium (and to a lesser extent, chromium), mapping the location, and indicating their role in heat-affected zone cracking.

Grain Size

It is generally conceded that heat-affected zone cracking tendency is minimized by avoiding large grains in the steels prior to fabrication. Early observations, however, are confusing in that some heats of Type 347 stainless steel having large grains performed reasonably satisfactorily, whereas finer grain-sized steels performed poorly. It is probable in these cases that other effects predominate. In other words, all other conditions being equal, it is still preferable to use small grain-sized steels for welded fabrication.

Blumberg, in a discussion of the paper by Soldan (Ref. 5), urges the use of fine grain steels, suggesting the beneficial additions of cerium which produce smaller grain-sized castings. Examining heat-affected zones in a fillet weld joining a coarse to a fine grained Type 347 stainless steel, Emerson (Ref. 42) found cracks only on the coarse grained heat-affected zone. In his Adams Lecture on the heat-affected zone, Nippes (Ref. 90) pointed out that because of the greater total grain boundary area in fine grained steels, whatever the cause of liquation cracks, the fine grained steels, being capable of distributing more of the offending constituent, should perform better than coarse-grained steels. On the other hand, Donati (Ref. 68) found the liquation temperature of boron treated Type 321 stainless steel to be unaffected by a difference in the grain size, and cited the work of Egnell (Ref. 91) who also found no difference in weldability of Type 321 stainless steel with variations in ASTM grain size from 1 to 4.

Robinson (Ref. 7), in his review article on liquation cracking, summarizes as follows: "Smaller grain size reduces liquation cracking, perhaps primarily by increasing grain boundary area. In parent steels, this may be controlled only during manufacture and may tip the balance between a relatively easily weldable material and one giving problems."

Back Filling

Evidence of penetration of the fused weld metal into the grain boundaries of the heat-affected zone has been reported by several investigators. On the theory that filling of grain boundaries would be more readily accomplished if the weld fusion zone were of a lower melting point than the base metal, Younger *et al.* (Ref. 19) chose to weld a crack-sensitive Type 347 stainless steel with a cobalt-base hard surfacing alloy having a melting point well below that of the stainless steel. No cracks were observed during the welding operation; on the other hand, with an austenitic stainless steel welding electrode, fine cracks adjacent to the fusion line were found.

Savage (Ref. 53) divided the heat-affected zone into two regions, which he defined as the partially melted zone and the "true" heat-affected zone. In these terms, then, the observations of Younger (Refs. 19, 20) were of the partially melted zone.

Moorhead and colleagues (Refs. 64, 74) point to backfilling in Type 304 stainless steel modified with small amounts of columbium. In Type 347 stainless steels, however, a wider partially melted zone exists and back filling cannot prevent cracking. They go on to explain, "We think this partially melted zone accommodates the strains [during Varestraint testing] produced by bending and solidification so that this region does not crack. Cracking does occur in the true HAZ (as in 347), but as there is no path for molten metal from the fusion zone to reach these cracks, backfilling cannot occur. . . . Apparently HAZ cracks can be backfilled only in those systems containing a very narrow partially melted zone. Thus, wide partially melted zones are detrimental to weldment cracking behavior."

Lippold (Ref. 75) also observed backfilling in alloy 800.

Conclusion

The integrity of thick austenitic stainless steel weldments to perform without heat-affected zone cracking over the intended service life is not yet assured by the controls which industry now applies. While the problems of the 1950's with Type 347 stainless steel are no longer the plague that they were then, they still persist. Cracks in Type 316 stainless steel, which was thought to be immune, are showing up in the 1980's, although in most cases only after twenty years of service at 1100°F (593°C) temperature. Meanwhile the attention of investigators has been focused on closely related alloys, such as Incoloy 800 and some of the nickel-base "superalloys" intended for elevated temperature service.

The causes and cures are not yet well understood. As the alloys become more complex, interactions must be faced, some of which may be encountered in the welding thermal cycle, or during postweld heat treatment. More often than not, the elusive causes are those which may result in failures only after years of high-temperature service.

What then have we learned about heat-affected zone cracking over the past thirty-five years? And what does this suggest should be studied to improve our understanding? Is the effort necessary and cost-effective? These are the questions which are dealt with in this concluding section.

Control of Cracking

Liquation cracks result from low melting eutectics which form on heating during the weld thermal cycle. Columbium and titanium carbides, CbC and TiC, form a eutectic with the austenite matrix, along with other constituents, such as Fe_2Cb, which is thought to be especially harmful in promoting liquation cracks. The initial conclusion, therefore, is that columbium and titanium-bearing austenitic stainless steels should be avoided.

To reach such a conclusion for avoiding problems in the weld heat-affected zone would cause the elimination of alloys deemed desirable for their creep-resisting properties. Since many austenitic stainless steels containing columbium and titanium have performed well in service, it would seem advantageous to learn more precisely the mechanisms which operate to cause failures. In this way the beneficial qualities imparted by these elements, such as creep resistance, could be employed.

Reheat or ductility dip cracking also contributes to heat-affected zone defects when columbium and, to a lesser extent, titanium are present. It has been demonstrated that the cause is related to the presence of fine intragranular carbides which strengthen the grains causing grain boundary decohesion under strain relaxation during postweld heat treatments or in service at elevated temperatures.

Precipitation is accelerated in the presence of plastic strains. Thus, for postweld heat treatments, a reduction of the peak welding stresses before reaching the temperature at which the maximum rate of carbide formation occurs is accomplished by holding at a temperature of 1100°F (593°C), prior to rapid heating to the solution annealing temperature of 1925°F (1052°C). By holding the weldment at 1650°F (899°C) following the solution anneal, the columbium carbides are stabilized and rendered inoffensive for long term service at elevated temperatures.

It is not yet clear whether upper limits on the amount of columbium or titanium would allow retaining the creep-resisting properties of Types 347 and 321 stainless steels and avoid the heat-affected zone cracking problem. This has been suggested in recent work on this subject and deserves further study. If undertaken, the question of the other elements which have been identified as contributing to heat-affected zone cracking problems should be considered. Most studies point to sulfur as a strong contributor to liquation cracks. Some would place phosphorus in the same category, although others believe that it is relatively harmless if kept within the normal limits; furthermore, phosphorus is beneficial for its contribution to creep properties.

Residual elements, including rare earth additions, although frequently studied by the alloy producers when considering their effects on hot formability, are insufficiently understood with respect to their contribution to heat-affected zone cracking. Likewise, the alloy balance with respect to the presence of ferrite in the microstructure of the steel before welding or the tendency to form ferrite in the heat-affected zone during the weld heating cycle needs further consideration. Interactions of the various elements during the thermal cycles or in long term service are subjects which deserve greater study.

Grain size control and homogenization are clearly important. This subject is especially pertinent when considering thick section weldments. Hot rolling and forging practices and subsequent heat treatments should desireably provide grain sizes not greater than ASTM 3 or even 4, if possible. Thick sections must be given hot forging or rolling reductions sufficient to eliminate ingot segregation and produce a homogeneous microstructure. This practice is not available to the foundries which have little possibilities to produce other than coarse ingot structures. Fortunately for them, most of the common austenitic stainless steels have small amounts of ferrite, which is believed to be desireable in welding thick section castings.

The choice of welding process and the control of the welding variables appear to have only a second order effect on the heat-affected zone cracking problem. Since liquation or precipitation effects are time and temperature dependent, it would be presumed that the welding procedures might have some influence. It seems to be desireable to keep the weld heating cycle as rapid as possible when dealing with crack sensitive alloys, suggesting the benefit of electron beam welding which has solved problems in the complex superalloys.

Arc welding procedure variables, though having only minor effects, should be restricted to those with low heat input. This should not exclude high energy processes, such as submerged arc for thick section welds, providing reasonably fast travel rates are employed with relatively small weld passes.

Preheat and interpass temperatures are generally conceded to have little effect, but most consider that they should be held down. There is evidence to suggest that very high preheats and interpass temperatures may be necessary in extreme cases, but then only if it is possible to hold them above the temperature at which the harmful precipitates will form. In Type 347 stainless steel, this has been suggested to be above 1382°F (750°C), which is normally impracticable.

Because the plastic strains in the heat-affected zone adjacent to thick section welds contribute to the cracking phenomenon, choices of filler metals which accept the cooling stresses by plastic deformation in the fusion zone will be beneficial to the heat-affected zone. The 16-8-2 filler metals have been used successfully for both Types 347 and 316 stainless steel weldments for the past thirty years. It should be noted that precise composition control of this filler metal may be required in order to provide the benefits attributed to it.

Inspection of weldments for heat-affected zone cracking during and after fabrication presents a difficult problem. This is because the cracks are generally microfissures and are not readily detected unless plastically deformed. Destructive test specimens prepared during welding procedure qualifications will generally identify the problem. Postweld radiographic procedures are usually ineffective. If the cracks are large enough, dye penetrant inspection will identify cracks which reach the surface. Ultrasonic techniques are currently being developed that will identify subsurface heat-affected zone defects. However, considerable skill is required to interpret the results, especially in stainless steels where the fusion zone has large dendritic grains giving confusing signals.

Directions for Further Research

Although this problem appears to have received considerable attention over the past thirty years, there remain questions which need systematic study. Some of these are:

1. Should the composition of austenitic stainless steels be more tightly controlled with respect to the principal alloying elements, the residual elements, and the trace elements?

2. Can the beneficial elements contributing to the creep strength of austenitic stainless steels be controlled within closer limits when they are found to contribute to heat-affected zone cracking?

3. What tolerances can be accepted due to through-thickness property variations, such as those caused by lack of sufficient hot working or those existing in castings?

4. What interactions exist during the various thermal cycles and how do these contribute to the problem?

5. With the recently available electron microscopic tools, can we not identify

more explicitly the failure mechanism relating to the formation of microconstituents?

6. Is it possible to assess the crack growth rate reliably enough to predict the remaining life of a weldment in elevated temperature service?

7. What are the possible environmental factors which contribute to heat-affected zone cracking and how can these be accommodated by changes in steel composition, protecting shields, stress alteration, etc.?

The benefits of research to solve this problem would seem to be enormous when considering the future requirement for thick section austenitic stainless steels and other related alloys. Some designers are now veering away from stainless steels or accepting lower design temperatures for their operation because of fears of encountering cracked heat-affected zones. Because thick sections of expensive alloys are most often relatively costly fabrications and because a failure may be costly in down time and in repair or replacement, the investment in research to solve the problem would appear to be a reasonable economic decision.

The users of thick section weldments, like the electric utilities and the chemical and petroleum process industries, should be significant beneficiaries of research leading to solutions to this problem. If it could reaffirm the use of Type 347 stainless steel with its higher stress allowables than the existing Type 316 stainless steel, thinner wall sections could be specified with attendant economic benefits in original cost, offering lighter support structures and greater flexibility of the piping systems in electric generating stations. Likewise, minor modifications of Type 316 stainless steel might well further improve its resistance to heat-affected zone defects and permit increases in minor elements, such as nitrogen and phosphorus, to give creep properties equivalent to Type 347 stainless steel.

Alloy producers and the fabricators have an obvious stake in research to solve this perplexing problem. The steel producer might overcome the risks of a rejected heat by becoming aware of the needed control of unwanted elements. Likewise, fabricators can overcome fears of schedule interruptions encountered with stainless steel heats which cause cracks during welding or postweld heat treatment.

Finally, the public gains economically from research in this field, as it does in many other fields, by lower costs of the products it purchases. It is, therefore, an important public policy that greater efforts be made to raise the level of funding for research where such economic benefits are apparent.

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References

1. ASTM Specification A376 and A430, Supplement 7.

2. Poole, L. K. 1953. The incidence of cracking in welding Type 347 steels. *Welding Journal* 42 (8): 403-s to 412-s.

3. Fleischmann, W. L. 1954. Determination of ferrite in Type 347 stainless steel weld deposits. *Welding Journal* 43 (9): 459-s to 468-s.

4. Curran, R. M., and Rankin, A. M. 1955. Welding of Type 347 stainless steel for 1100°F turbine operation. *Welding Journal* 44 (3): 205-213.

5. Soldan, H. M., and Mayne, C. R. 1957. Ductility related to service performance of heavy-wall austenitic pipe welds. *Welding Journal* 46 (3): 141-s to 147-s.

6. Gooch, T. G., Howard, R. D., McKeown, D., and Widgery, D. J. 1975. Assessment of the weldability of high temperature austenitic materials. *Proceedings international conference on welding research related to power plant,* Southampton, 17-21 September, CEGB publication, paper 29: 401-411.

7. Robinson, J. L., and Scott, H. M. 1980. Liquation cracking during the welding of austenitic stainless steels and nickel alloys. *Proceedings conference, residuals, additives and materials properties,* London, 15-17 May, 1978. Royal Society publication ISBN 0-85403-132-4: 105-117.

8. Puzak, P. P., Apblett, W. R., and Pellini, W. S. 1956. Hot cracking of stainless steel weldments. *Welding Journal* 45 (1): 9-s to 17-s.

9. Manjoine, M. J. 1976. *Crack initiation and growth in welds at 1100°F (595°C).* Westport, Connecticut: Society for Experimental Stress Analysis.

10. Lundin, C. D., and Spond, D. F. 1976. The nature and morphology of fissures in austenitic stainless steel weld metals. *Welding Journal* 55 (11): 356-s to 367-s.

11. Hemsworth, B., Boniszewski, T., and Eaton, N. 1969. Classification and definition of high-temperature welding cracks in alloys. *Metal Construction* 1 (2S): 5-16.

12. Kerr, J. G. Private communication.

13. Biemiussa, K., Grosser, E. D., Lorenz, H., and Ziebig, H. 1974. Design and requirements for SNR pipework. *Nuclear Engineering International* 19 (213): 94-96.

14. Edmonds, D. P. Private communication.

15. Whitaker, J. S., and Jessop, T. J. 1981. Ultrasonic detection and measurement of defects in stainless steel—a literature survey. *British Journal of Non-Destructive Testing* 23 (6): 293-303.

16. Haddrill, D. M., and Baker, R. G. 1968. HAZ cracking in thick section austenitic steels during postweld heat treatment. *BWRA Bulletin* 6(11): 282-288.

17. Engineering assessment of an advanced pulverized-coal power plant. 1982. EPRI reports CS-2223 & CS-2555 (projects 1403-1 & 1403-2).

18. DeLong, J. S. Private communication.

19. Younger, R. N., and Baker, R. G. 1960. Heat-affected zone cracking in welded high-temperature austenitic steels. *Journal of the Iron and Steel Institute* 196(2): 188-194.

20. Younger, R. N., and Baker, R. G. 1961. Heat-affected zone cracking in welded austenitic steels during heat treatment. *British Welding Journal* 8(12): 579-587.

21. Irvine, K. J., Murray, J. D., and Pickering, F. B. 1960. The effect of heat-treatment and microstructure on the high-temperature ductility of 18% Cr-12% Ni-1% Nb steels. *Journal of the Iron and Steel Institute* 196(2): 166-179.

22. Private communication.

23. Kapranos, P. A., and Whittaker, V. N. 1982. Ultrasonic inspection of fatigue cracks in austenitic 316 and 347 weldments. *British Journal of Non-Destructive Testing* 24(3): 129-134.

24. Nippes, E. F., and Savage, W. F. 1949. Development of specimen simulating weld heat-affected zones. *Welding Journal* 28(11): 534-s to 546-s.

25. Nippes, E. F., Savage, W. F., Bastian, B. J., Mason, H. F., and Curran, R. M. 1955. An investigation of the hot ductility of high temperature alloys. *Welding Journal* 34(4): 183-s to 196-s.

26. Savage, W. F., and Lundin, C. D. 1965. The Varestraint test, *Welding Journal* 30(10): 433-s to 442-s.

27. Savage, W. F., and Lundin, C. D. 1966. The application of the Varestraint test to the study of weldability. *Welding Journal* 31(11): 497-s to 503-s.

28. Goodwin, G. M. PhD thesis. Troy, New York: Rensselaer Polytechnic Institute.

29. Truman, R. J., and Kirby, H. W. 1960. Some ductility aspects of 18-12-1Nb steel. *Journal of the Iron and Steel Institute* 196(2): 180-187.

30. Christoffel, R. J. 1960. Notch-rupture strength of Type 347 heat-affected zone. *Welding Journal* 39(7): 315-s to 320-s.

31. Pickard, A. C., Ritchie, R. O., and Knott, J. F. 1975. Fatigue crack propagation in a Type 316 stainless steel weldment. *Metals Technology* 2(6): 253-263.

32. Corum, J. M., and McAfee, W. J. 1982. Creep rupture, fatigue, and creep-fatigue of weldments. *Presentation to subgroup on elevated-temperature design.* November 10, 1982.

33. Younger, R. N., Borland, J. C., and Baker, R. G. 1961. Heat-affected zone cracking of two austenitic steels during welding. *British Welding Journal* 8 (12): 575-578.

34. Sorokin, L. I., and Elkin, I. S. 1975. A wedge butt test for assessing the weldability of

cast steels and alloys. *Automatic Welding* 28 (8): 4-6.

35. Gooch, T. G., and Honeycombe, J. 1975. Microcracking in fully austenitic stainless steel weld metal. *Metal Construction* 7 (3): 146-148.

36. Haddrill, D. M., and Baker, R. G. 1965. Microcracking in austenitic weld metal. *British Welding Journal* 12 (8): 411-419.

37. Gordine, J. 1971. Some problems in welding Inconel 718. *Welding Journal* 50 (11): 480-s to 484-s.

38. Arata, Y., Terai, K., Nagai, H., Shimizu, S., and Aota, T. 1977. Fundamentale Studien zum Electronenstrahlschweissen von Hitzebeständigen Legierungen für Kernkraftanlagen (Bericht 3) — Metallurgische Betrachtungen über Mikroriss, *Transactions of the Japanese Welding Research Institute* 6 (2): 235-250.

39. Brooks, J. A. 1974. Effect of alloy modifications on HAZ cracking of A-286 stainless steel. *Welding Journal* 53 (11): 517-s to 523-s.

40. Brooks, J. A., and Krenzer, R. W. 1974. Progress toward a more weldable A-286. *Welding Journal* 53 (6): 242-s to 245-s.

41. Wylie, R. W. 1958. Cooperative investigation of a new welding electrode for stainless steel. *Welding Journal* 37 (9): 426-s to 432-s.

42. Emerson, R. W., and Jackson, R. W. 1957. The plastic ductility of austenitic piping containing welded joints at 1200°F. *Welding Journal* 36 (2): 89-s to 104-s.

43. Van Bemst, A. 1983. *Postweld heat treatment of stainless steels and nickel alloys in thick sections,* International Institute of Welding document IX-1288-83.

44. Christoffel, R. W. 1962. Cracking in Type 347 heat-affected zone during stress relaxation. *Welding Journal* 41 (6): 251-s to 256-s.

45. Naiki, T., Okabayashi, H., Kuribayashi, M., and Morishige, N. Cracking in welded 18 Cr-12Ni-Nb steel during stress relieving (in Japanese). *Ishikawajima-Harima Giho* 15: 209-215.

46. Morishige, N., Kuribayashi, M., Okabayashi, H., and Naiki, T. On the prevention of service failure in Type 347 stainless steel. Third international symposium of the Japan Welding Society. Paper no. 3. JWS-65.

47. Younger, R. N., Haddrill, D. N., and Baker, R. G. 1963. Postweld heat treatment of high-temperature austenitic steels. *Journal of the Iron and Steel Institute* 201 (8): 693-698.

48. Christoffel, R. J. 1963. Notch sensitivity of the heat-affected zone in Type 316 material. *Welding Journal* 42 (1): 25-s to 28-s.

49. Caughey, R. H., and Benz, Jr., W. G. 1960. Material selection and fabrication, main steam piping for Eddystone no. 1, 1200°F and 5000 psi service. *Journal of Engineering for Power* 82 (4): 293-314.

50. Johnson, W. C., and Blakeley, J. M., eds. 1979. *Interfacial Segregation.* ASM Seminar on Interfacial Segregation, Chicago, 1977. American Society for Metals.

51. Tamura, H., and Watanabe, T. 1973. Mechanism of liquation cracking in weld heat-affected zone of austenitic stainless steel. *Trans. Japan Welding Society* 4 (1): 30-42.

52. Tamura, H., and Watanabe, T. 1972. Study on the behaviour of grain boundary at heat-affected zone on SUS 347 and SUS 321 austenitic stainless steels. *Journal of the Japan Welding Society* 41 (9): 1094-1108 (in Japanese); also *Welding Research Abroad* 20 (6): 38-50 (1974).

53. Savage, W. F. 1980 Houdremont lecture: solidification, segregation and weld imperfections. *Welding in the World* 18 (5-6): 89-114: International Institute of Welding.

54. Guttmann, M., and McLean, D. 1979. Grain boundary segregation in multicomponent systems. *Interfacial Segregation:* 261-348: Metals Park, Ohio: American Society for Metals.

55. Cullen, T. M., and Freeman, J. W. 1963. Metallurgical factors influencing hot ductility of austenitic piping at weld heat-affected zone temperatures. *Journal of Engineering for Power* 85: 151-164.

56. Arata, Y., Matsuda, F., and Saruwatari, S. 1974. Varestraint test for solidification crack susceptibility in weld metal of austenitic stainless steels. *Trans. Japan Welding Society* 3 (1): 79-88.

57. Honeycombe, J., and Gooch, T. G. 1973. Effect of microcracks on mechanical properties of austenitic stainless-steel weldmetals. *Metal Construction Suppl.* 5 (4): 140-147.

58. Sadowski, E. P. 1974. Modification of cast 25Cr-20Ni for improved crack resistance. *Welding Journal* 53 (2): 49-s to 58-s.

59. Morishige, N., Kuribayashi, M., and Okabashi, H. 1979. *Effects of chemical composition of base metal on susceptibility to hot cracking in austenitic stainless steel welds.* IIW document IX-1114-79.

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

61. Swindeman, R. W., Sikka, V. K., and Klueh, R. L. 1983. Residual and trace element effects on the high-temperature creep strength of austenitic stainless steels. *Metallurgical Transactions A* 14: 581-593.

62. Goodall, P. J., Cullen, T. M., and Freeman, J. W. 1967. The influence of nitrogen and certain other elements on the creep-rupture properties of wholly austenitic Type 304 steel. *Trans. ASME Journal of Basic Engineering* 89: 517-524.

63. Lai, J. K. L. 1981. Optimizing the stress rupture properties of AISI 316 stainless steel with the aid of computers. *Journal of Nuclear Materials* 99: 148-153.

64. Moorhead, A. J., and Sikka, V. K. 1979. Effect of residual niobium on Type 304 stainless steel. *Welding Journal* 58 (9): 253-s to 261-s.

65. Loveless, D. L., and Bloom, F. K. 1957. Boron solves "hot shortness" in stainless steels. *The Iron Age* 179 (25): 95-97.

66. Goldschmidt, H. J. 1971. Effect of boron additions to austenitic stainless steels — part I: effect of boron additions to 20%Cr, 25%Ni austenitic steel with and without Nb, Mn, Si; Part II: solubility of boron in 18%Cr, 15%Ni austenitic steel. *Journal of the Iron and Steel Institute* 209 (11): 900-911.

67. Brooks, J. A., Thompson, A. W., and Williams, J. C. 1980. Weld cracking of austenitic stainless steels — effects of impurities and minor elements. *Proceedings of Symposium on Physical Metallurgy of Joining:* Metallurgical Society of AIME, ISBN 0-89520-365-0:117-136.

68. Donati, J. R., Spiteri, P., and Zacharie, G. 1973. Etude de la fissuration à chaud des zones affectées par le soudage d'aciers austénitiques du Type 18-10 stabilisés au titane: influence de la teneur en bore. *Mémoires Scientifiques de la Revue de Métallurgie* 70 (12): 939-950.

69. Donati, J. R., Guttmann, D., and Zacharie, G. 1974. Influence de la teneur en bore sur la tendence à la fissuration à chaud dans les zones affectées par le soudage d'acier inoxydables 18-10. *Revue de Métallurgie* 71 (12): 917-929.

70. Boudot, R., and Zacharie, G. 1982. Influence de la teneur en bore sur la resistance à la fissuration à chaud dans les zones affectées par le soudage d'acier 18%Cr-12%Ni au molybdène en relation avec le mode d'élaboration. *25ème, Colloque de Métallurgie — Progrès Récents dans l'Elaboration des Métaux et Alliage, Conséquences sur leur Propriétés d'Emploi:* Saclay, 23-25 Juin 1982.

71. Williams, T. M., Harries, D. R., and Furnival, J. 1972. Creep-rupture behaviour of Type 316 austenitic steel in relation to boron content. *Journal of the Iron and Steel Institute* 210 (5): 351-358.

72. Edmonds, D. P., King, R. T., and Goodwin, G. M. 1979. Residual elements have significant effects on the elevated-temperature properties of austenitic stainless steel welds. *Proceedings: symposium of austenitic stainless steels and their weld metals: Atlanta, 14 November 1977; STP 679:* 56-68: Philadelphia, Pennsylvania; American Society for Testing Materials.

73. Sikka, V. K., Morehead, A. J., and Brinkman, C. R. 1979. Influence of small amounts of niobium on the mechanical and corrosion properties of Type 304 stainless steel. *Ibid:* 69-102.

74. Moorhead, A. J., Sikka, V. K., and Reed, R. W. 1979. Effect of small additions of niobium on the welding behavior of an austenitic stainless steel. *Ibid:* 103-123.

75. Lippold, J. C. 1983. An investigation of heat-affected zone cracking in alloy 800. *Welding Journal* 62 (1): 1-s to 11-s.

76. Persson, N. G. 1971. The influence of sulphur on the structure and weldability of a titanium-bearing austenitic stainless steel. Proceedings of the Soviet-Swedish Symposium. *Clean Steel* 1: 142-151: Sandviken, Sweden.

77. Mayer, G., and Clark, C. A. 1974. A review of the known effects of trace elements in steels and alloys containing nickel. *Metallurgist and Materials Technologist* 6 (11): 491-501.

78. Pruger, T. A., Blake, F., and Valley, J. A. 1967. The effects of residual elements on the cold rolling and finishing of austenitic stainless steels. *STP 418:* 24-36. Philadelphia, Pennsylvania, American Society for Testing Materials.

79. Nakamura, Y., and Abe, S. 1977. The manufacture and characteristics of ultra-high-purity austenitic stainless steels by the MSR process. *Nippon Steel technical report overseas no. 10.*

80. Hull, F. C. 1960. Effects of alloying additions on hot cracking of austenitic chromium-nickel stainless steels. *Proceedings ASTM* 60: 667-690.

81. Matthews, S. J., Maddock, M. O., and Savage, W. F. 1972. How copper surface contamination affects weldability of cobalt superalloys. *Welding Journal* 51 (5): 326-s to 328-s.

82. Savage, W. F., Nippes, E. F., and Mushala, M. C. 1978. Copper contamination cracking in the weld heat-affected zone. *Welding Journal* 57 (5): 145-s to 152-s.

83. Savage, W. F., Nippes, E. F., and Stanton, R. P. 1978. Intergranular attack of alloy

steels by molten copper. *Welding Journal* 57 (1): 534-s to 546-s.

84. Nippes, E. F. 1980. Copper-contamination cracking in the heat-affected zone of welds. *Proceedings of symposium: physical metallurgy of metal joining, St. Louis, Mo., 16-17 October 1980: Metallurgical Society of the AIME.* ISBM 0-89520-365-0: 53-73.

85. Nippes, E. F., Wawrousek, H., and Fleischmann, W. L. 1955. The heat-affected zone in arc-welded type 347 stainless steel. *Welding Journal* 34 (4): 169-s to 179-s.

86. Nippes, E. F., Schaaf, B., Fleischmann, W. L., and Mehan, R. L. 1957. Time-temperature effect on properties of weld heat-affected zone in Type 347 stainless steel. *Welding Journal* 36 (6): 256-s to 270-s.

87. Morishige, N., Kuribayashi, M., and Okabayashi, H. 1982. Study of weldable 18Cr-10Ni-Nb stainless steels. *IHI Engineering Review* 15 (2): 1-6.

88. Bloom, F. K., Clarke, Jr., W. C., and Jennings, P. A. 1951. Relation of structure of stainless steel to hot ductility. *Metal Progress* 59 (2): 250-256.

89. Donati, J. R., Lehuede, G., and Zacharie, G. 1975. Influence de la tendence ferritique sur la susceptibilité à la fissuration à chaud au cours du soudage d'aciers austénitique 18-10. *Mémoires Scientifiques Revue Métallurgie* 72 (11): 823-836.

90. Nippes, E. F. 1959. The weld heat-affected zone. *Welding Journal* 38 (1): 1-s to 18-s.

91. Egnell, L. 1970. Welding trials on a titanium-bearing austenitic steel. *Welding Institute Conference on Welding of Creep Resistant Steels:* paper 12. Newcastle, England.