Weldability and Solidification Phenomena of Cast Stainless Steel

The weldability of cast alloy CF-8M is found to be largely dependent on its initial solidification mode, with primary ferrite heats exhibiting greater weldability

BY M. J. CIESLAK AND W. F. SAVAGE

Introduction

Stainless steel casting alloys have been employed successfully for many years in the severe environments encountered in the chemical, petroleum, and nuclear industries. An entire class of alloys has been developed for use in corrosive environments at working temperatures below 1200°F (649°C). Unfortunately, in many cases where acceptable mechanical properties and corrosion resistance have been achieved, both castability and weldability have been poor. Since there is often a correlation between castability and weldability, a better understanding of the solidification mechanics of these materials should make the design of improved alloys possible.

Solidification research based on commercial alloy systems is difficult. This is both because of the complexity of the alloys and because the high temperatures involved make direct observation of the solidification process difficult. As a result, most information on the solidification mechanics of metal systems has been based upon room temperature metallographic examination of the solidification substructures of castings and weldments and upon mathematical theories.

In castings, the cooling times are normally long enough for diffusion-controlled processes to modify both the microstructure and pattern of microsegregation during cooling from the solidus to room temperature. Conversely, the cooling times for most weldments are so short as to suppress most diffusion-controlled reactions in the solid state. Thus, the microstructure and the pattern of microsegregation established during solidification of weld metal is far better preserved. Therefore, a more accurate interpretation of the mechanics of solidification can be obtained from examination of the microstructure and the pattern of microsegregation in weld metal at room temperature.

The redistribution of solute during solidification has a strong influence on both the weldability and the castability of many stainless steels. "Hot tears" in castings and "hot cracks" in welds appear to occur by similar mechanisms as a direct result of the microsegregation of certain solute elements.

Hot Cracking Studies

Studies of the solidification of austenitic stainless steels have been conducted for many years. It has been observed since the early 1940's that hot cracking in austenitic stainless steel weldments is reduced or prevented when a small amount of delta ferrite is present. Early studies by Borland and Hull supported the fact that 5-10 vol-% of delta ferrite in weldments and castings afforded the best resistance to hot cracking and hot tearing. Since that time, a voluminous amount of data which is in general agreement with these initial observations has been generated and reported by other investigators.

Several theories have been suggested to explain weld metal hot cracking. Early work by Bochvar, Pumphrey and Jennings, and Lyons, resulted in the formulation of the Shrinkage-Brittleness Theory. This theory postulates a semi-continuous thin liquid film which is susceptible to hot cracking if shrinkage strains exceed some critical level. The theory also states that, since diffusion is limited in welding, microsegregation is augmented and the resulting low melting constituents aggravate the hot cracking problem. This theory also recognizes the possibility of "healing" or "backfilling" of cracks with low melting point constituents, such as eutectic liquids, if they are present in large enough quantities.

A second theory—the Strain Theory of Hot Cracking—offered by Pellini and Appleby proposes that a continuous, thin liquid film present at some stage of the solidification process cannot support the strains induced in the weldment or casting without hot cracking.

The Generalized Theory of Super-Solidus Cracking as proposed by Borland is probably the most widely accepted theory of hot cracking and hot tearing today. Borland postulates that a semi-continuous liquid film exists in the weldment or casting separated by bridges of solidified material. The solid bridges, when subjected to a critical level of induced strain, will fracture, forming hot cracks. Borland uses a composition-dependent surface energy argument in conjunction with rapid solidification rates to describe
the form and composition of the last-to-solidify material.

In addition to theories postulating a liquid phase as a prerequisite for hot cracking, a solid-state theory of hot cracking has been proposed by Movchan. Several tests have been developed to rate the relative hot cracking tendencies of commercial austenitic stainless steels. These include the Fis Dickson Test,\(^3\) the Varestraint Test,\(^3\) the Cast Pin Tear Test,\(^3\) the Sure Bend Test,\(^3\) and many others. These tests generally conclude that austenitic stainless steels containing a few volume-percent delta ferrite offer a much higher resistance to hot cracking than do wholly austenitic stainless steels.

All fusion welding processes involve microsegregation of the component alloying elements. In stainless steel weldments, chromium (Cr) is rejected to the liquid during primary austenite solidification and nickel (Ni) is rejected to the liquid during primary delta ferrite solidification. Elements such as silicon (Si), molybdenum (Mo), sulfur (S), and phosphorus (P), which are more soluble in delta ferrite, are rejected to the liquid during primary austenite solidification. Elements such as carbon (C) and manganese (Mn) are more soluble in austenite and are thus rejected to the liquid during primary delta ferrite solidification.

Sulfur and phosphorus form low melting eutectics between sulfides and phosphides, and silicon tends to form glassy silicate films. In alloys solidifying as primary austenite, sulfur, phosphorus, and silicon are rejected to the liquid, form low melting constituents in the interdendritic regions during the terminal transient period, and thus cause hot cracking. Alloys solidifying as primary delta ferrite reject fewer of these elements to the liquid and consequently are more resistant to hot cracking.

![Solidus and liquidus curves of the Fe-Cr-Ni system](image.png)

**Table 1—Chemical Compositions and Ferrite Contents of Seven CF-8M Heats**

<table>
<thead>
<tr>
<th>Heat no.</th>
<th>C</th>
<th>Mn</th>
<th>Si</th>
<th>Cr</th>
<th>Ni</th>
<th>Mo</th>
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<td>.048</td>
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Figure 1 consists of a projection of the solidus and liquidus curves on a plane of the Fe-Cr-Ni ternary diagram.\(^21\) According to Fig. 1, compositions on the nickel-rich side of the dashed liquidus curve should solidify as primary austenite while compositions on the chromium-rich side should solidify as primary delta ferrite. Masumoto et al.\(^22\) reported that welds solidifying on the nickel-rich side of the liquidus were extremely susceptible to hot cracking whereas welds solidifying on the chromium side did not hot crack.

The DeLong\(^8\) and Schaeffler\(^6\) diagrams attempt to predict room temperature ferrite contents from the compositions of the alloys. Predicting weldability based on these diagrams has met with mixed success. Masumoto\(^23\) found no correlation between weldability and ferrite content based on these diagrams. Recent work by Lippold\(^6\) has shown that the form and volume of room temperature ferrite is a function of composition, the shape of the Fe-Cr-Ni phase diagram, and the cooling rate experienced by the material.

**Materials and Procedures**

Materials for this investigation were supplied by member foundries of the Steel Founder's Society of America. Seven heats of CF-8M containing varying amounts of delta ferrite were received. All heats were annealed for one hour at 2050°F (1121°C) and water quenched. The chemical compositions of the heats are given in Table 1. These compositions were determined by wet analysis and are the average of three analyses. It should be noted that the carbon content of heat 6 and the nickel content of heat 3 were greater than allowed by CF-8M specifications.

Ferrite contents as measured by point counts taken of the as-cast microstructure in accordance with ASTM E-562 are listed in Table 1. Magne-gage readings for both the as-cast and weld microstructures are also.
listed. Test specimens were machined from selected positions in a 12 in. (305 mm) long keel block casting as shown schematically in Figs. 2 and 3.

The Varestraint Test was used to determine the weldability of the 7 heats received. Prior to testing, all specimens were cleaned with acetone and demagnetized with an ac de-gaussing coil.

The welding parameters for this investigation are given in Tables 2 and 3. In order to maintain an elliptical weld pool approximately ½ in. (12.7 mm) wide, voltages and currents were adjusted as shown in Table 3 for each heat.

Following Varestraint testing, the as-welded surface was examined under a stereo-microscope equipped with a filar eyepiece. At each level of augmented strain, the total crack length was determined for each specimen by summing the lengths of the individual cracks. The maximum crack length and number of cracks at each strain level were also determined.

Metallographic specimens were removed from the test bars as shown in Fig. 4. A section parallel to the weld surface, as well as longitudinal and transverse cross-sections of the weld table.

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**Table 2—Constant Parameters**

| Electrode composition: tungsten + 2% thorium | Electrode: conical, 90 deg included angle | Electrode extension from collet ~ 1/16 in. (16.7 mm) |
| Electrode diameter: 5/32 in. (3.18 mm) | Arc length ~ 3/16 in. (2.4 mm) (measured cold) | Travel speed: 4 ipm (1.7 mm/s) |
| Gas flow rate: 40 cfm argon (18.9 L/min) | Current: DCSP |

**Table 3—Variable Parameters**

<table>
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<tr>
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</tr>
</tbody>
</table>
were prepared. Specimens were mounted in epoxy and polished through 0.06 micron alumina. Etchants used included Marble's reagent, glycerigia, a mixed acid etch, and a 10% oxalic acid electroetch. The caption associated with each photomicrograph indicates which etchant was employed.

Electron microprobe specimens were mounted in 1 in. (25.4 mm) epoxy-filled graphite shells suitable for placement in the specimen holder of a Materials Analysis Company 400 microprobe. These were also polished through 0.06 micron alumina and lightly etched to reveal both phase boundaries and backfilled hot cracks. Microhardness indentations were then introduced to bracket specific areas for investigation with the microprobe.

A beam diameter of approximately 2 microns was used throughout the investigation. A 20 kV accelerating potential and a specimen current of approximately 0.25 mA were used throughout. The point count method of analysis was employed. Ten-second counts were repeated at least twice for each point to reduce the possibility of electrical error. Raw data were reduced to weight-percentages using the MAGIC IV computer algorithm developed by Colby.

Results and Discussion

Varestraint Results

Figure 5 shows the total crack length vs. percent augmented strain relationships of the seven heats. Where additional specimens were available, duplicate tests were run to substantiate the data.

Several heats of CF-8M were found to be extremely resistant to hot cracking. These heats experienced no visible hot cracks on the as-welded surface up to an augmented strain of 2%. The major difference among individual CF-8M heats was the amount of room temperature ferrite present in the as-cast microstructure.

The relationship between the average total crack length with 2% augmented strain and the volume-percent ferrite in the room temperature cast microstructure is summarized in Fig. 6.
The data indicate that the sensitivity to hot cracking tends to decrease with increasing volume-percent ferrite. However, the lack of consistency in the trend suggests that other factors are involved.

Analysis of Microstructure

Pseudo-binary diagrams, consisting of sections of the Fe-Cr-Ni ternary at constant iron content, are helpful in understanding the microstructure of these alloys. However, it should be emphasized that pseudo-binary diagrams can only provide qualitative information relative to the solidification of commercial alloys, because the synergistic effects of carbon, molybdenum, manganese and silicon on solidification phenomena are poorly understood. Therefore, the pseudo-binary approach will be used only as an aid in visualizing primary effects rather than as a basis for making quantitative predictions.

The Fe-Cr-Ni system is complicated by the fact that while the binary Fe-Cr system exhibits a minimum at 1516°C (2761°F) and 21% Cr, the binary Fe-Ni system exhibits a peritectic at 1517°C (2763°F) and 4.3% Ni. The Fe-Cr-Ni system exhibits a ternary eutectic at < 1300°C (2372°F) and 9 Fe-48 Cr-43 Ni. Thus, as the chromium level is increased, alloys in the iron-rich corner of the ternary system experience a transition from peritectic to eutectic behavior. Reference to Fig. 1 reveals that this transition occurs at the point where the dashed liquidus crosses the solidus curve at the 90% iron line.

Thus, alloys which contain 90% or more iron and lie between the dashed liquidus and the righthand solidus exhibit peritectics. Alloys containing less than 90% iron which fall between the solidus curves exhibit ternary eutectics.

Comparison of Figs. 7 and 8 illustrates this fact.

In Fig. 7, the pseudo-binary section at 93% Fe, the liquidus is located to the Ni-rich side of both solidus lines to form a typical skewed peritectic triangle. On the other hand, in Fig. 8, with the pseudo-binary section at 70% Fe, the liquidus lies between the two solidus lines as is typical of a ternary eutectic triangle. Since commercial stainless steels contain significantly less than 90% Fe, it seems likely that the invariant reaction associated with the terminal transient period of solidification will be an eutectic and the pseudo-binary will resemble that shown in Fig. 8.

Figure 9 shows such a pseudo-binary diagram in schematic form with the approximate relative locations of heats 1 through 7 indicated by the numbered arrows.
Within the triangular eutectic region, a three-phase equi­brium exists between austenite, delta ferrite, and liquid. Alloys having compositions on the nickel-rich side of the apex of the three-phase eutectic triangle solidify as primary austenite. Alloys having compositions on the chromium-rich side solidify as primary delta ferrite. In either case, during the terminal transient period, the last remaining liquid is of eutectic composition. However, since the amount of eutectic liquid is extremely small, the interdendritic regions consist of a divorced eutectic.

In alloys which solidify as primary austenite, the divorced phase at the grain and subgrain boundaries is austenite. The ferrite which forms during the initial transient period for alloys that solidify as primary ferrite is sufficiently enriched in chromium and depleted in nickel to be stable as ferrite at room temperature. Thus, the cores of the subgrains remain as delta ferrite while the remainder of the subgrains experience a transformation to austenite. By contrast, in alloys which solidify as primary austenite, the divorced eutectic ferrite that is retained is located at the grain and subgrain boundaries.

According to their estimated location on Fig. 9, heats 4-7 would be expected to solidify as primary delta ferrite, as has been confirmed by metallographic observation.

Heats 4-6 exhibit the vermicular delta ferrite structure characteristic of weldments with ferrite numbers between 4 and 20. The views of Fig. 10 are representative of this vermicular delta ferrite morphology. Inspection of Varestraint results for these materials shows that heats 4 and 5 are highly resistant to hot cracking, which is in agreement with the usual theories. However, heat 6 shows anomalous behavior. Inspection of Table 1 indicates that the levels of sulfur and phosphorus in this heat were the highest of the seven heats tested in this study. Microprobe analysis across a backfilled crack in this material, shown at X250 in Fig. 11A, revealed a significant increase in phosphorus over the matrix levels. Apparently, the nominal phosphorus content of this heat is too large to be partitioned effectively to the primary delta ferrite, and the resulting segregation pattern lowers the melting temperature in intergranular regions.

The cracking which occurred in the heat 6 material was located exclusively at grain boundaries as no intercellular...
cracking was found. Metallographic examination revealed that the vermicular ferrite structure was absent in the grain boundaries of heat 6. Microprobe analysis detected an increase in nickel and a decrease in chromium at the grain boundaries; this is in agreement with the postulate that the last material to solidify in this heat would be austenite.

Figure 11B summarizes the results of the microprobe analysis made of the backfilled crack shown in Fig. 11A. No difference in molybdenum contents could be detected between the backfilled region and the matrix. The behavior of this heat indicates that the presence of primary δ-ferrite will not necessarily ensure good weldability. Control of the residual elements known to have a deleterious effect on the weldability of stainless steels is also important.

Heat 7 also solidified as primary delta ferrite. The microstructure, shown at X250 in Fig. 12, indicates that the composition of this alloy lies in a chromium-rich region far to the right of the three-phase triangle. Primary δ-ferrite solidification consumed almost the entire liquid, with austenite forming only at the final intergranular sites. During cooling to room temperature, much of the primary ferrite transformed to austenite by a Widmanstätten-type nucleation and growth mechanism. Inspection of Figure 12 reveals that the austenite formed at the grain boundary provides nucleation sites from which the dark etching Widmanstätten austenite plates grow into the ferrite grains. Inspection of Varestraint results for this heat shows that heat 7, which also solidified as primary ferrite, is highly resistant to hot cracking.

Heats 1 to 3 solidified as primary austenite. Inspection of Varestraint data indicates that the three heats are highly susceptible to hot cracking. In theory, ferrite should form in the final interdendritic and intergranular solidification sites. Lippold found this eutectic ferrite in backfilled cracks in Type 310 stainless steel Varestraint specimens and reported a Cr/Ni ratio of approximately 2.3 in this region.

David et al. report finding no interdendritic ferrite using a magnetic etching technique in interrupted solidification samples of Type 310 stainless steel. They also found no indication of the formation of an eutectic ferrite phase with the differential thermal analysis techniques used in their studies. However, they used a cooling rate of 6°C/min (11°F/min), which is at least three orders of magnitude slower than those found in normal GMA, GTA, and SMA weldments. It is also questionable if differential thermal analysis can resolve the crystallization of a second phase when less than 1 vol-% of this phase is formed during the solidification process.
Finally, Lippold\textsuperscript{5} has shown that the nominal composition within the primary austenite region greatly influences the amount of eutectic ferrite formed. The closer the composition is to the ternary liquidus, the greater is the amount of eutectic ferrite formed. This result agrees qualitatively with Fleming's analysis\textsuperscript{38} of a similar situation in binary Al-Cu alloys. The fraction of eutectic is an exponential function of the ratio between the nominal alloy composition and the eutectic composition.

The compositions of commercially available Type 310 stainless steels are sufficiently removed from the appropriate ternary liquidus points to ensure that only a very small amount of ternary eutectic may form.

Figure 13 clearly establishes the existence of eutectic ferrite at final solidification sites between primary austenite cellular dendrites and in intergranular volumes in heats 1, 2, and 3.

Close inspection of Fig. 13C, taken of heat 3 at X250, reveals an almost continuous interdendritic eutectic ferrite network to be present in this weldment. Microprobe analysis was performed across the eutectic ferrite shown in Fig. 14A. The results, summarized in Fig. 14B, indicate the ferrite to be enriched in chromium and molybdenum and depleted in nickel. Note that the molybdenum content of the ferrite is more than three times that of the adjacent austenite. The effect of molybdenum as a ferritizer has been well documented,\textsuperscript{29,30,39-40} but usually with respect to its influence on the ferrite content of materials which solidify with ferrite as the primary phase. Undoubtedly, the partitioning of molybdenum to interdendritic volumes greatly aids in ferrite formation.

Hot cracking in primary austenite alloys has been related to many factors, including the partitioning of harmful ferritizing elements to final solidification sites. Views B and C of Fig. 15 are images of the microstructure of the backfilled crack shown in Fig. 15A made with the microprobe using characteristic X-rays. Figure 15B was made with phosphorus Ka radiation. Figure 15C represents a summation of the sulfur Ka and molybdenum La which are almost identical in wavelength. Phosphorus segregation to this region is clearly evident in Fig. 15B. Although the segregation shown in Fig. 15C is obviously greater, the sulfur Ka peak is so close to the Mo La peak that it is not possible to obtain a quantitative estimate of the amount of sulfur present in the backfilled crack.

Cast Structures

Inspection of the microstructure found in the as-cast CF-8M alloy revealed that more ferrite was present than would be expected in comparable wrought materials, such as Type 316 stainless steel. The segregation in castings is on a far coarser scale, and the solidification substructure size are several orders of magnitude larger.
Fig. 16—Cast and annealed microstructures of Heats 1-7, (A-G, respectively) all electro-etched.

Ferrite in the primary austenite heats is in the form of a divorced eutectic, rather than primary, ferrite.

Comparison of Magne-gage readings for heats 3 and 4 illustrates the danger of using this type of ferrite determination to predict weldability. The difference in ferrite numbers (3.0 vs. 2.6) does not correlate with the weldability as revealed by the Varestraint test results shown in Fig. 6. Thus, comparison of weld solidification substructures must be made before the weldability can be predicted accurately.

Massive Transformation

Lippold has recently postulated that the conversion of primary delta ferrite to austenite could occur by a diffusionless massive transformation as a result of the rapid cooling through the two-phase delta-plus-gamma region. This would apply to alloy compositions which lie on the chromium-rich side of the three-phase triangle on the pseudo-binary diagram where the primary solid phase is delta ferrite.

Figure 17 compares ferrite morphologies at the fusion boundary of autogenous welds in alloys which solidified as primary delta ferrite and as primary austenite, respectively. The ferrite-austenite interfaces in the primary austenite heats exhibited none of the angular facets which would be expected with a massive transformation.

Solidification as primary delta ferrite results in chromium segregation at the spines of the ferrite dendrites. Thus, even with slow cooling, retained ferrite has a higher chromium content than the austenite. Figure 18 shows that for a nominal alloy composition $C_0$, primary dendrite spines should have a composition approaching $C_F$. In fact, it is possible for regions within the ferrite to have any composition between $C_F$ and $C_0$.

The liquidus temperature for this cast structure will have a range determined by the slopes of the solidus and liquidus and the nominal composition of the alloy. The result is that the ferrite islands in the casting will have a higher melting temperature than the surrounding matrix. When the alloy is welded, it is probable that ferrite
islands in the base metal at the solid-liquid interface do not melt, while adjacent austenite of roughly nominal composition does melt.

The ferrite has an immense temperature gradient impressed upon it as a result of its intimate contact with the liquid on one side of the island and the heat sink provided by the solid base plate on the other side. This rapid cooling of ferrite through the two-phase austenite-plus-ferrite field permits the occurrence of the massive delta to gamma-phase transformation previously mentioned. Inspection of the ferrite islands in Fig. 17A shows the angular crystallographic facets characteristic of some massively transformed materials.

Recent STEM microanalysis by Lyman has shown a compositional variation of less than 1 wt-% of chromium and nickel across the massive transformation interface. STEM micrographs indicate a wall of dislocations located at the interface, providing a mechanism for the diffusionless shear transformation.

Conclusions

1. The results of Varestraint testing support the observations that stainless steels which solidify with delta ferrite as the primary phase are less susceptible to hot cracking than those in which austenite is the primary phase.

2. Segregation of phosphorus to final solidification sites has been shown to occur in heats in which austenite is the primary phase, thus supporting the hypothesis that this element is partitioned to the liquid during primary austenite solidification.

3. Phosphorus may also be detrimental to heats which solidify as primary delta ferrite if its concentration is in excess of that which may be effectively absorbed by the ferrite.

4. Since the CF-8M alloy composition range appears to straddle the ternary liquidus curve, either ferrite or austenite may serve as the primary phase, depending on the actual composition.

5. Divorced eutectic ferrite is observed at final interdendritic and intergranular solidification sites when austenite is the primary phase.

6. Microanalysis reveals this ferrite to be enriched in both molybdenum and chromium.

7. Ferrite is present in the as-cast and annealed base metal structure in amounts greater than would normally be found in wrought materials of similar composition, regardless of the primary solidification mode.

8. Because of the coarseness of the as-cast and the annealed microstructures, it is not possible to identify whether ferrite or austenite was the primary phase from the ferrite morphology.

9. Magne-gage analyses alone do not appear to be sufficient to determine whether or not a specific heat of material is readily weldable.

10. The angular, faceted ferrite morphology observed at the fusion boundary in heats in which ferrite is the primary phase resembles that associated with massive diffusionless
transformation found in other alloy systems.
11. Analysis of available ternary Fe-Cr-Ni phase diagrams leads to a conclusion that an eutectic reaction is the invariant reaction characterizing the end of solidification in commercial stainless steels.

Acknowledgments

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References

11. Pellini, W. S., Foundry, 80(11), 1952.
28. Colby, J. W., MAGIC IV, Bell Telephone Laboratories, Allentown, PA.
33. Lyman, C. E., unpublished research at Rensselaer Polytechnic Institute.