Technical Note: Ferrite Stability at Elevated Temperature in Austenitic Stainless Steel Weld Metal

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When austenitic stainless steel weld metals are exposed to elevated temperatures during service or postweld heat treatment, the delta ferrite constituent in the duplex austenite-ferrite microstructure tends to transform to the sigma (σ) and, to a lesser extent, chi (χ) intermetallic phases. Carbides, particularly $M_{23}C_6$, are also formed especially during the early stages of aging. The presence of these phases in the microstructure causes the creep rupture, tensile ductility and impact properties to be reduced compared to the as-deposited property levels. This effect is, of course, dependent on the distribution and amount of the ferrite present initially and the extent of transformation to sigma and other phases.

The need for a duplex ferrite-austenite microstructure in the as-deposited weld metal has arisen from the desire to produce fissure-free deposits. The mechanism by which weld metal compositions producing duplex as-deposited microstructures inhibit fissure formation has been studied extensively and the available information recently reviewed by DeLong[1] and Lundin and Spond.[2] However, the specific mechanisms remain undefined at this time.

The fissuring tendency of various austenitic stainless steel weld metals as a function of ferrite level was rigorously documented by Lundin, DeLong, and Spond.[3] As a result of this study, numerous undiluted SMA weld deposits with varying ferrite levels were available for use in aging studies. By utilizing specimens extracted from these weld pads, aging studies on Types 308, 308L, 316, and 316L stainless steel weld metals were conducted. The compositions of the deposited weld metals are given in the literature,[3] and the ferrite levels designated by the A, B, C, and D denotations indicate nominal aim ferrite numbers of 0, 2, 4, and 6 respectively.

After the initial ferrite levels of the extracted as-deposited specimens were determined, the specimens were aged in an inert atmosphere for times varying upward to 2,000 h. The results of the aging studies at 550 and 650 C (1022 and 1202 F) for the Type 316 deposits are shown in Fig. 1. The four as-deposited levels—A, B, C, and D—are shown at time 0 at the left in Fig. 1A and B.

The influence of temperature on the rate and extent of transformation after various times is clearly evident. At 550 C the ferrite has fallen to half its original level in 200 h, while at 650 C one-half of the original level is reached in 20 h. It is also to be noted that the ferrite is completely transformed after 200 to 2,000 h at 650 C. The Type 316L behaved in a manner similar to 316, and the 308 aging behavior was virtually identical to that of the 308L. Thus, the difference in carbon content within each grade did not appear to influence the decomposition or transformation rate of ferrite.

If the results of the aging studies are plotted such as to reflect the fraction

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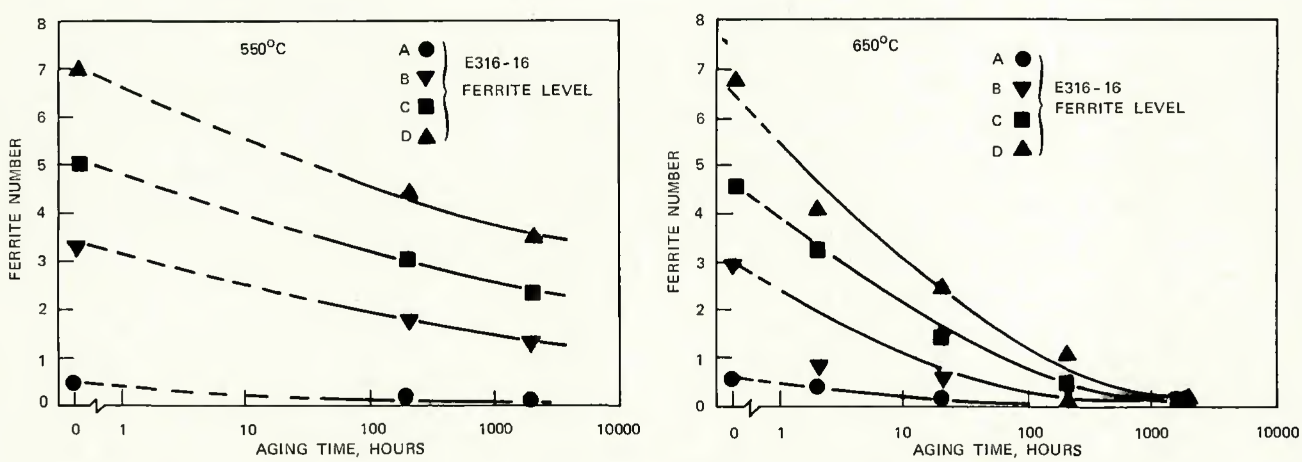

Fig. 1—Ferrite number vs. aging time—Type 316 stainless steel weld metal: A (left)—550 C; B (right)—650 C

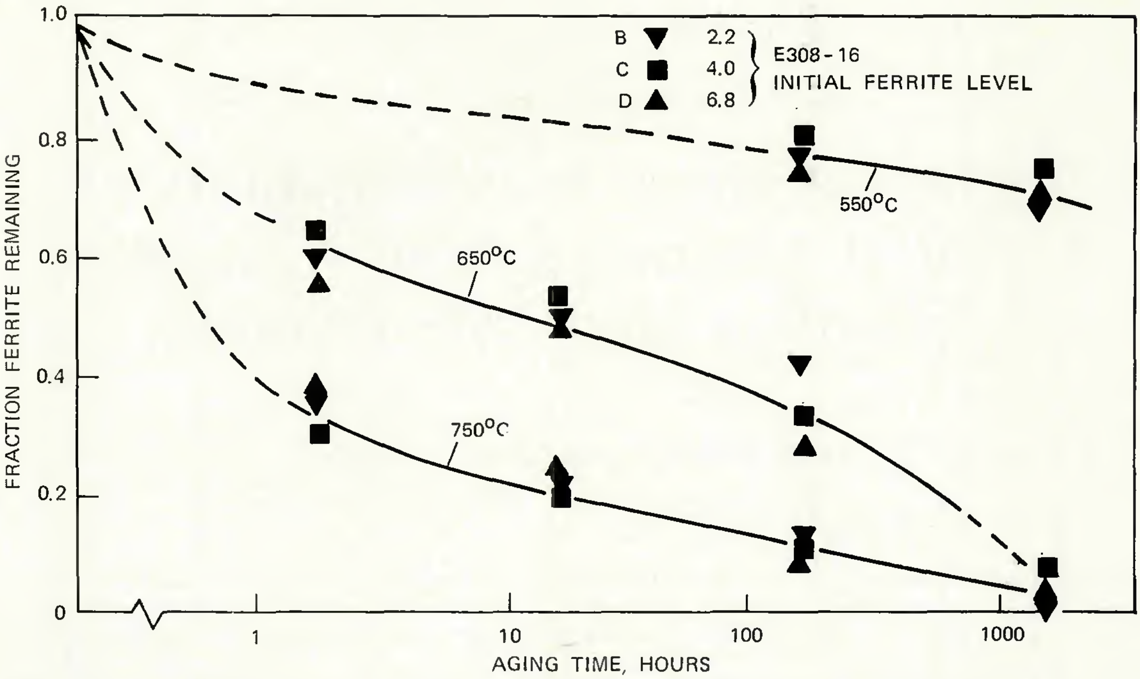

Fig. 2—Fraction of ferrite remaining as a function of aging time for Type 308 stainless steel weld metal

of ferrite remaining as a function of time, the initial ferrite level can be eliminated as a parameter. This type of presentation is shown in Fig. 2 for Type 308 stainless steel weld metal at the three aging temperatures utilized. It is clear that the initial ferrite level has little effect on its rate of decomposition for 308 weld metal. Similar results were obtained for 308L, 316, and 316L weld metals.

While there appears to be some scatter in the data of Fig. 2, the scatter after aging is no more than that which results from ferrite number determinations alone, regardless of the pre-treatment. When the 316 and 316L weld metals were compared on this basis, the rate of austenite decomposition was greater than that for the 308 and 308L types. For example, the fraction of ferrite remaining in 316 or 316L after 200 h at 650 C is approximately 0.15 while that for 308 and 308L types is approximately 0.40. The greater decomposition rate for the 316 and 316L types may be due to the greater chromium equivalent (as determined by Hull's calculation[4]) when compared to the 308 and 308L compositions.

The phases formed in the aged weld metals were identified by X-ray diffraction from extracted residues (10% HCl in methanol electrolyte). The major constituent extracted from the samples aged at 550 C for 2,000 h was $M_{23}C_6$. The samples aged at 650 C and 750 C (1202 and 1382 F) contained large quantities of sigma as well as some $M_{23}C_6$.

Microstructural examination of the aged samples showed that, when the ferrite number exceeded approximately 6 and the aging time and temperature were sufficient for complete transformation, the sigma network was continuous throughout the weld metal substructure boundaries. The brittle network so produced can have pronounced influence on the mechanical properties. This influence on mechanical properties will be further documented by a continuation of the present work at the University of Tennessee, a study currently sponsored by the Metal Properties Council at Battelle-Columbus, and studies at Oak Ridge National Laboratory, Metals and Ceramics Division. Further evidence from the metallographic studies indicates that the sigma phase may continue to grow after all of the initial ferrite phase has been consumed in the transformation process. This appears to be more prevalent in the 316 and 316L materials with their higher chromium equivalents.

The aged samples produced in this study will be used in a companion study aimed at determining the corrosion behavior of weld metals with varying austenite-ferrite-sigma contents. This investigation will be reported at a later date. The authors acknowledge the Stainless Steel Advisory Subcommittee of the Welding Research Council for making the previously manufactured weld pads available for study.

References

1. DeLong, W. T., "Ferrite in Austenitic Stainless Steel Weld Metal," 1974 Adams Lecture: *Welding Journal,* 53 (7), July 1974, Res. Suppl., pp. 273-s to 286-s.

2. Lundin, C. D., and Spond, D. F., "The Nature and Morphology of Fissuring in Austenitic Stainless Steel Weld Metals," *Welding Journal,* 55 (11), Nov. 1976, Res. Suppl., pp. 356-s to 367-s.

3. Lundin, C. D., DeLong, W. T., and Spond, D. F., "Ferrite-Fissuring Relationship in Austenitic Stainless Steel Weld Metals," *Welding Journal,* 54 (8), Aug. 1975, Res. Suppl., 241-s to 246-s.

4. Hull, C. F., "Effects of Composition on Embrittlement of Austenitic Stainless Steels," *Welding Journal,* 52 (3), March 1973, Res. Suppl., pp. 104-s to 113-s.