



## Hydrogen-Induced Cracking along the Fusion Boundary of Dissimilar Metal Welds

*The susceptibility of dissimilar austenitic/ferritic combinations to hydrogen-induced cracking near the fusion boundary has been investigated*

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**ABSTRACT.** Presented here are the results from a series of experiments in which dissimilar metal welds were made using the gas tungsten arc welding process with pure argon or argon-6% hydrogen shielding gas. The objective was to determine if cracking near the fusion boundary of dissimilar metal welds could be caused by hydrogen absorbed during welding and to characterize the microstructures in which cracking occurred. Welds consisted of ER308 and ER309LSi austenitic stainless steel and ERNiCr-3 nickel-based filler metals deposited on A36 steel base metal. Cracking was observed in welds made with all three filler metals. A ferrofluid color metallography technique revealed that cracking was confined to regions in the weld metal containing martensite. Microhardness indentations indicated that martensitic regions in which cracking occurred had hardness values from 400 to 550 HV. Cracks did not extend into bulk weld metal with hardness less than 350 HV. Martensite formed near the fusion boundary in all three filler metals due to regions of locally increased base metal dilution.

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### Introduction

Dissimilar metal welds are used extensively in the power generation, petrochemical and heavy fabrication industries. Numerous instances of cracking along the dissimilar metal fusion boundary have been reported, particularly in cladding applications where a corrosion-resistant austenitic alloy is applied to a ferritic structural steel. Often this cracking, or disbonding, has been associated with exposure to hydrogen in service and, as a result, the mechanism has been described by various authors as a form of hydrogen-induced cracking (Refs. 1-13). This type of cracking has been reproduced in the laboratory by exposing austenitic cladding to hydrogen, either in an autoclave or by cathodic charging (Refs. 1-3, 7, 8, 11-13).

In practice, however, this form of

cracking has occurred during fabrication, prior to exposure to a hydrogen environment. The fact that disbonding can occur without prolonged exposure to hydrogen in service suggests that either hydrogen is not necessary for disbonding to occur, or hydrogen absorbed during welding can cause cracking near the dissimilar metal fusion boundary.

The fusion boundary microstructure in dissimilar welds often possesses some unique features. Normal epitaxial nucleation during solidification along the fusion boundary gives rise to grain boundaries that are continuous from the base metal into weld metal across the fusion boundary. These boundaries are roughly perpendicular to the fusion boundary and have been referred to as "Type I" boundaries. In dissimilar welds, where an austenitic weld metal and ferritic base metal exist, a second type of boundary that runs roughly parallel to the fusion boundary is often observed. This has been referred to as a "Type II" boundary (Ref. 6). These boundaries typically have no continuity across the fusion boundary to grain boundaries in the base metal. Several investigators have reported that hydrogen-induced disbonding typically follows Type II grain boundaries (Refs. 1-4, 7, 8, 12,13). The disbonding phenomenon that occurs following fabrication and prior to service has also been associated with these Type II boundaries.

An additional complication in austenitic/ferritic dissimilar welds is the dramatic transition in composition and

### KEY WORDS

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Austenitic Stainless  
Filler Metals  
Nickel-Based Filler  
GTAW  
Martensite







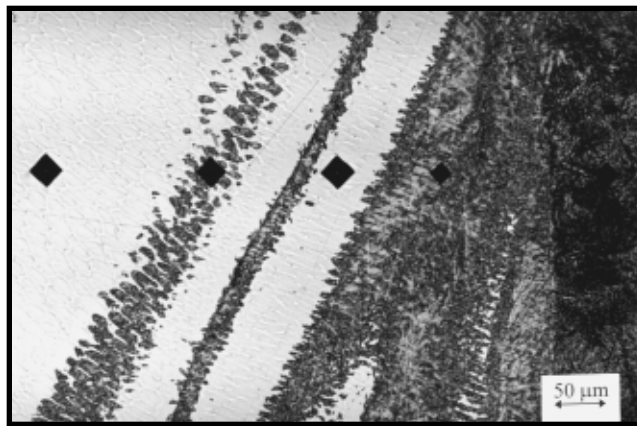
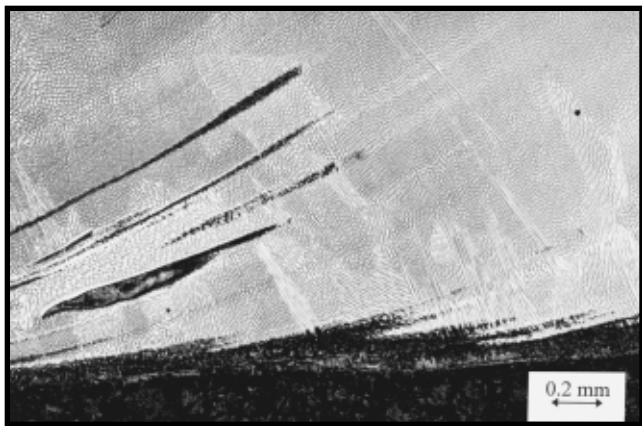


Fig. 7 — Plan view, ER309LSi filler metal, longitudinal strain, 33% dilution, pure argon shielding gas, chromic acid/nital/ferrofluid. A — Microstructure of mostly austenite with bands of martensite visible; B — bands of martensite near the fusion boundary.

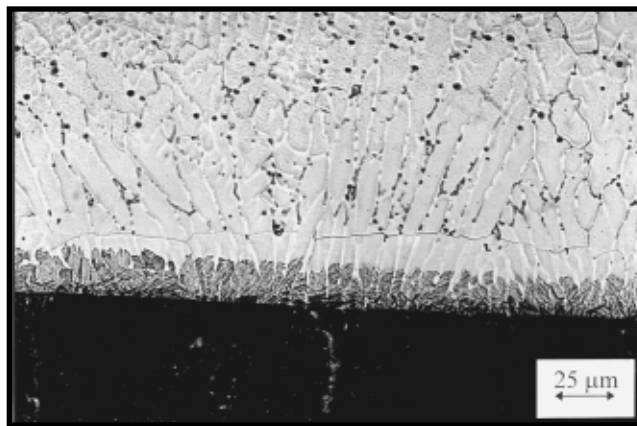
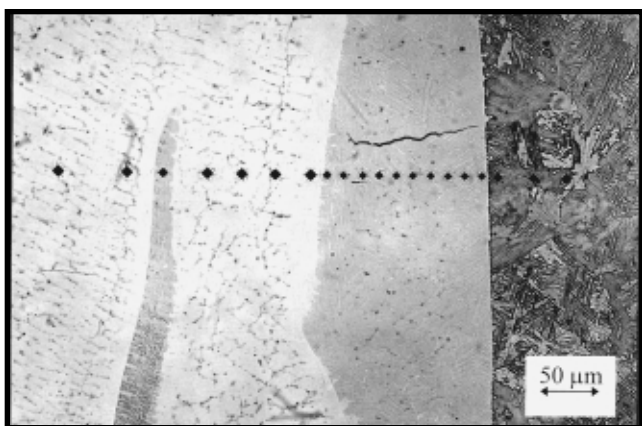


Fig. 8 — Plan view, ER309LSi filler metal, longitudinal strain, 38% dilution, Ar-6% $H_2$  shielding gas, chromic acid/nital etch showing a martensitic region (dark gray) near the fusion boundary containing a microscopic crack.

Fig. 9 — Plan view, same weld as Fig. 8, ferric chloride electrolytic etched revealed a martensitic microstructure near the fusion boundary and Type II grain boundaries.

Fig. 2 are associated with light-etching bands in the weld metal. Figure 3 shows cracks propagating parallel to the fusion boundary in a sample subjected to transverse strain. Figure 4 shows the microstructure of the weld, which appears in Fig. 2 as revealed by ferrofluid applied to the polished surface. The region near the fusion boundary where the cracking occurred is strongly colored by ferrofluid (blue or purple) in Fig. 4A and has an average hardness of 476 HV, which is significantly higher than the bulk weld metal average hardness of 311 HV.

The region near the center of Fig. 4B labeled A+F has a microstructure of austenite plus skeletal ferrite, which is typical of ER308 weld deposits. The austenite is white or light brown while the ferrite is colored. The region labeled A+M has a microstructure of austenite plus martensite. Hardness indentations in Fig. 4B indicate that the region colored

by ferrofluid has a significantly higher hardness than the white region, which is consistent with the presence of martensite. When compared to Fig. 2, it was found that the light etching bands where cracking initiated corresponded to the A+M regions in Fig. 4. It can be concluded from the high hardness and coloration by ferrofluid that the cracking occurs in regions containing martensite.

#### ER309LSi Weld Deposits

Type 309LSi filler metal exhibited less severe cracking than Type 308 for a given dilution. Figure 5 shows cracking in a region of high hardness near the fusion boundary in a weld made with Ar-6% $H_2$  shielding gas. The crack extends 600  $\mu m$  away from the fusion boundary. Figure 5B shows the propagation path at the tip of the crack in Fig. 5A. Both intergranular and transgranular propagation are ev-

ident. This is consistent with Savage, *et al.* (Ref. 17), who observed both intergranular and transgranular propagation in experiments involving direct observation of hydrogen-induced cracking. The chromic acid etch does not clearly reveal the austenite plus martensite structure through which the crack propagates. The average hardness in the crack region, as determined from the traverse shown in Fig. 5A, is 431 HV, while the bulk weld metal hardness is 239 HV.

Figure 6 shows the same weld that appears in Fig. 5 treated with ferrofluid after being etched with chromic acid and nital. Martensitic regions are brown while austenitic regions remain white. Hardness indentations indicate a significantly higher hardness in the regions colored by ferrofluid, which is consistent with the presence of martensite. Cracking is visible in the band of high hardness martensite in the weld metal adjacent to



