ABSTRACT. The formation of a thin, hard and brittle layer at the interface between the weld and the base metal during the gas metal arc welding process was studied for the case of a high-strength low alloy steel welded with a stainless steel electrode. This embrittlement is usually attributed to diffusion across the dissimilar metal interface, whereas the results of the present study point to segregation as the more likely cause. The proposed segregation model, which is based on the mechanism of solidification, shows a better fit with the experimental results obtained.

Introduction

“Dissimilar metal welding,” viz., one in which a filler metal different from the base metal is used, is a common practice. In the welding of low alloy, high-strength armor steels, an austenitic filler metal is used extensively, because the preheating is less critical than that required when welding with a ferritic electrode.

The deposition of stainless, austenitic filler metal on a low alloy steel gives rise to several effects that may have a significant influence on the quality of the weldment.

The first of these effects is well understood and results from the dilution of the filler metal caused by the local melting of the base metal and its mixing with the weld metal. In ordinary welding conditions, dilution up to 40% may occur, depending on the welding technique used.

Another effect observed (Ref. 1-4) is the formation of a distinct intermedium region at the interface between the weld metal and the heat-affected zone. The appearance of this region has usually been attributed to interdiffusion of the alloying elements in the system formed between the base metal and the weld metal system. It is generally accepted that this intermediate zone is of a martensitic structure and that its chemical composition ranges between the composition of the weld and the composition of the base metal (Ref. 1, 3, 4).

The diffusion model assumes migration of the Ni atoms of alloying elements of the weld metal into the base metal (Ref. 3, 4) and diffusion of carbon atoms from the base metal into the weld (Ref. 2). This diffusion process arises from the composition difference at the weld metal interface and the high-temperature cycles accompanying the welding.

Seferian (Ref. 2) has shown that carbon diffusion occurs between carbon steels and heat-resistant alloyed Cr-Mo or austenitic steels exposed for a long time to high temperatures. Diffusion of carbon across dissimilar metal joints has been studied by Eckel (Ref. 5). He studied the diffusion of carbon atoms across the interface between a carbon steel and an austenitic stainless steel. At 1300°F (704°C) the diffusion was slow and 23 hours (h) were required to form a decarburization band ~225 μm thick. Ikawa et al. (Ref. 3) and others (Ref. 1, 4) explained the formation of the intermediate zone as being the result of the diffusion of Cr and Ni atoms across the weld/base metal interface at high temperatures simulated by the thermal cycle of welding. Ikawa et al. (Ref. 3) calculated the Ni atoms distribution, assuming diffusion through the weld/base metal interface. Using the following expression:

\[ C(x,t) - C_i = \frac{1}{2} \left( C_f - C_i \right) \left[ 1 - \text{erf} \left( \frac{x}{2\sqrt{D}t} \right) \right] \]

where \( C(x,t) \) is the Ni composition distribution, \( C_f \) is the Ni content in the base metal, and \( C_i \) is the Ni content in the weld metal. The location of the fusion line is at \( x = 0 \).

Ikawa et al. (Ref. 3) also established that the formation of the intermediate martensite zone is reduced with increasing Ni content of the electrodes. When the Ni content of the electrodes reached ~60%, no intermediate zone was observed.

Another explanation for the formation of the boundary layer has been put forward by Cookson (Ref. 9). He suggested a mechanism whereby the first deposited droplets layer the molten crater bottom and spontaneously solidify to form a thin layer of highly diluted weld metal upon which are fused successive beads. However, this hypothesis is not compatible with the stirring that is known to occur before the solidification of the weld pool.

In the present paper it is shown that the formation of an intermediate zone is best explained on the basis of a model of segregation during the solidification process of the weld metal. This offers a better explanation than the diffusion model for the formation of the intermediate zone.

Materials and Experimental Procedure

Bead-on-plate tests were carried out by depositing a wire austenitic filler metal on low alloy high-strength steel. Gas metal arc (GMA) welding was used. The welding conditions are given in Table 1. The compositions of the filler metal and the base metal are given in Table 2.

The plates were cut perpendicular to the welding direction, and samples for metallographic investigations were removed. The samples were subsequently ground and polished, standard specimen preparation techniques being used. The specimens were etched with Barlie’s reagent (10 g CuSO₄, 50 ml HCl, 50 ml H₂O) in order to emphasize the intermediate zone formed between the weld and the base metal.

Microhardness scans across the boundary between the weld and the
base metal were carried out with a Vickers type microhardness tester. Chemical composition scans across the same boundary were made with the aid of the Proxan EPMA system attached to a scanning electron microscope.

Experimental Results

Metallographic examinations of the boundary area between the weld and the base metal showed a well defined band of darkly etched material located between the weld and the heat-affected zone (HAZ)—(Fig. 1). The microhardness in this region was found to be higher than either in the HAZ or in the weld (above 500 DPH, as compared with ~300 DPH in the weld and ~400 DPH in the HAZ). The thickness of the intermediate zone was found to be between 5 and 200 µm, depending on the heat input, the preheat conditions during welding and the location of the area under consideration as related to the position of the arc.

Figures 2 and 3 show typical scans of the elements Cr, Ni, Mn and Fe across weld/base metal boundary in welded specimens preheated to 200°C (392°F) and 400°C (752°F). The local content of the above elements is plotted as a function of distance across the weld. Two distinct distribution curves are noted in the intermediate zone (boundary layer, BL) changing gradually from the composition of the weld. The actual position of the fusion line (FL) is assumed to coincide with the composition of the metal begins to change (for all the elements). It is known (Ref. 10) that the micrographic contrast is dependant on the etching procedure; therefore, the dark boundary cannot be considered confidently as a fusion line. Our designation of FL fits consistently within the frame of our model for boundary formation as delineated below.

Discussion

Diffusion Model of Intermediate Boundary

The diffusion models proposed by Ikawa (Ref. 3) and others (Ref. 2, 4), which were briefly described in the Introduction, are the generally accepted explanations for the formation of the intermediate brittle zone during dissimilar-metal welding. However, an analysis of the diffusion models shows that they are not able to explain fully the formation of the intermediate zone and to define its thickness. Let us assume a composition profile for Ni as given by equation (1) by assuming that:

\[ C_{Ni}^t \ (x, t) - C_{Ni}^o = 0 \]  
and:

\[ \frac{\text{erf}(x/2)}{\sqrt{\pi}} \] 

and from tables took the argument to be equal to 2.4. This, however, is incorrect since erf \( \infty = 1 \) by definition. Therefore, subsequent relations, based on equations (3), proposed by Ikawa would also appear to be essentially erroneous.

The diffusion profile given by equation (1) can be used for estimating the thickness of the diffusion layer, if a Ni composition range for martensite formation is assumed. Introduction of boundary concentrations for martensite formation \( C_{Ni}^{\text{max}} \) and \( C_{Ni}^{\text{min}} \) into equation (1) permits the evaluation of the thickness of the martensitic layer, \( \Delta \):

\[ \Delta = x_{\text{max}} - x_{\text{min}} \]  
where \( x_{\text{max}} \) and \( x_{\text{min}} \) are locations of the Ni boundary concentrations in the transition zone.

The concentrations \( C_{Ni}^{\text{max}} \) and \( C_{Ni}^{\text{min}} \) can be roughly estimated from the Schaeffler Diagram. Let us assume, on the basis of this diagram, that boundary concentrations \( C_{Ni}^{\text{min}} = 2\% \) and \( C_{Ni}^{\text{max}} = 6\% \) (for dissimilar weld 18/8 stainless steel and pure iron). In this case, the Ni initial concentrations are \( C_{Ni}^o = 0 \) and \( C_{Ni}^i = 8\% \). Substituting these values into Equation (1) enables the thickness of the intermediate martensitic zone to be expressed as a function of the diffusion coefficient \( D \) and time \( t \): equation (5):

\[ \Delta = x_{\text{max}} - x_{\text{min}} = 1.88 \sqrt{D t} \]
Assuming a diffusion coefficient of Ni in iron at 1300°C, $D \approx 10^{-7}$ cm$^2$/second (s), the time required to form a layer $\Delta = \mu$m thick is $\sim 30$ s. This seems to be much too long for the diffusion models to be applicable in explaining the formation of the intermediate, martensitic zone.

One has to bear in mind that these evaluations are very much simplified. In this respect, the diffusion process, structural changes, etc., are not only influenced by the Ni content but also by the presence of other elements, which were not considered. Nevertheless, this does not diminish the justification of questioning the suitability of the diffusion models for explaining intermediate zone formation during dissimilar metal welding.

In addition to the anomalous diffusion time calculated, the form of the solute distribution curve between weld metal and matrix is not compatible with a distribution due to diffusion.

**Solidification with Segregation Model**

A better representation of all the data obtained can be achieved by considering that the formation of the intermediate zone is associated with the solidification process. The model presented below is based accordingly on an appropriate solidification theory.

Solidification may occur with no solid diffusion, limited liquid diffusion, and no convection. When it does, transient solute profile is formed in the liquid phase during the solidification process. Similar effects associated with some welding parameters—equation (8). Introducing, into that equation, $\delta$ and $k$ calculated for Ni (see Table 2) and the diffusion coefficient $D_k = 5.78 \times 10^{-7}$ cm$^2$/s (Ref. 8) enables the solidification rate $R$ at the boundary to be estimated. $R$-values established for experiments A and B are given in Table 3.

The rate of solidification was found to be much smaller than the welding velocity (0.36 cm/s). This difference between the two rates is not surprising, since at the weld boundary the solidification rate is always much lower than the welding velocity. At a certain temperature in the melting range the diffusion coefficient $D_k$ is constant; practically only two of the parameters in equation (8) change, viz., $R$ and $\delta$.

The solidification rate $R$ is dependent on the heat flow across the liquid/solid boundary and on the welding velocity. A high preheat results in the formation of a thick intermediate band because of the decrease of the solidification rate. Similar effects would be observed with the reduction in welding velocity because of a...
decrease in the solidification rate. In practice, however, the thickness of the boundary layer is influenced primarily by the mixing of the weld pool. The mixing is due to some arc effects and explains the observed wide variability of the thickness of the layer in the weld.

The microstructure of the boundary layer is directly related to its chemical composition, which changes gradually from that of the base metal to that of the deposited metal. Most of these compositions fall within the martensite range of the Schaeffler Diagram. This explains why this band exhibits such a high microhardness.

It can be seen from Figs. 2 and 3 that there is a slight difference in the composition profiles obtained in specimens preheated to 400 and 200°C (752 and 392°F). In specimens preheated to the higher temperature, some rounding up of the distribution profile was observed, while in specimens preheated to the lower temperature, the transition is abrupt. This effect is due to diffusion processes that seem to be more intensive at higher preheat temperatures during the welding cycle.

The range of applicability of the segregation model is limited to low values of the solute partition ratio k. When the weld metal is richer in alloying elements, the k ratio increases and the solidification pattern changes. It was reported (Ref. 3 and 4) that an increase of the Ni content in the weld metal causes a delay of the segregation band also, when the Ni content is high enough, the first solidifying phase is the γ-phase and not δ-ferrite, and the formation of the martensitic boundary layer is completely eliminated.

Conclusions

1. The formation of an intermediate layer of martensitic structure between the austenitic weld metal and the ferrite base metal is probably caused by transient segregation resulting from the solidification process, with limited diffusion in the liquid and no convection.

2. The solid diffusion of alloying elements across the weld boundary is a secondary effect and pertains to cases in which the joint is exposed for a long time to high temperatures.

3. The thickness of the boundary layer can be estimated roughly from the parameters characterizing the welding process.

References


