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EFFECT OF WELDING ON THE FRACTURE BEHAVIOUR OF THE FERRITIC CORROSION-RESISTANT STEEL

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Resume
Structural instability of the ferritic corrosion-resistant steel typically results in the carbides precipitation on the coarsened grain boundaries and in the associated intercrystalline corrosion. The examined steel revealed a tendency towards a defective fracture behaviour associated with the different carbide phase distribution. In the case of a preferential distribution within the grain interior, the relaxation processes along the welding line are limited and the transgranular cleavage fracture mechanism plays a role. Intergranular distribution of secondary phases brought tendency towards intergranular decohesion.


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1. Introduction

Ferritic and semiferritic non-stabilized steels find wide application not only because of the lower associated cost as compared to the stabilized or austenitic variants of anticorrosion steels. In fact, the more stable mechanical parameters at higher degrees of forming are an asset as well.

In welded structures, application of such steels is often limited by processes induced in the heat-affected zone (HAZ). These mostly include loss of passivation capabilities and typical intergranular corrosion due to grain coarsening and diffuse carbon redistribution. Enrichment of the grain boundaries with carbon and subsequent formation of chromium carbides are most pronounced in the zone of the highest heat effect, i.e. along the fusion line. The process is highly dependent on the steel's internal homogeneity, structure status, residual stress from the forming process, etc. [1]. The problem of local loss of corrosion resistance is frequently addressed also for austenitic corrosion-resistant steels [2, 3], e.g., by supplementary chemical passivation or by restricting the thermal power during welding [4]. For stabilized steels, weldability is improved (i.e. the grain coarsening process is limited and the structure is stabilized) by addition of Ti or by global microalloying with Ti and Nb [5]. However, stabilization with Ti has its limitations, particularly in highly oxidative environments, due to the preferential attack of highly oxidative acids on titanium carbides [6]. The consequences of destabilization of titanium carbides have also been observed during pickling after welding when using a HNO3 based pickling agent [7].

Initiation of intergranular damage also without apparent corrosion has been observed in non-stabilized ferritic steel 1.4016 in operating conditions. The present analyses were aimed at structural changes in the HAZ of this type of steel resulting in changes in the fracture behaviour. In dependence on the degree of heating in the various HAZ sub-layers, different material degradation mechanisms were observed, resulting in defective fracture in static tests.
2. Work methodology

This work was a continuation of the analyses of damage of heterogeneous welded joints in sections of material 1.4016 and austenitic steel 1.4301. The TIG welding method in a pure Ar protective atmosphere using 1.4370 type filler in the form of an austenitic welding rod (OK Tigrod 16.95, ESAB) was used.

The tendency towards intergranular corrosion was the major issue limiting the lifetime of the weldments in sea shore environments. In addition to that issue, a tendency towards intergranular damage and occasionally transgranular damage was observed in the HAZ of ferritic steel 1.4016. Crack initiation was also observed in regions beyond the reach of the effect of the typical carbon redistribution, in other words, beyond the reach of the degradation effect of the carbides formed along the ferritic grain boundaries.

The weldment parts damaged in this way were analyzed with a view to elucidating the damage mechanism and its relation to the observed structure changes in the HAZ. The fracture behaviour was studied based on fractographical analysis of the fracture planes following static load. For this, experimental samples of weldments were prepared, representing the above combination of the materials welded and enabling uniaxial loading. They were samples of sheets 1.5 mm thick, 20 mm wide and 120 mm long. Structural changes in the HAZ sub-layers were examined on metallographic sections prepared perpendicularly to the fusion zone. The formation of secondary phases, observed in relation to the change in the fracture mode, was evaluated by using energy dispersive microanalysis. The structural changes influence on the mechanical parameters was evaluated by the hardness measurement.

3. Achieved results

3.1 Structural changes resulting from welding

The initial structural state of the investigated steel 1.4016 is shown in Fig.1; its chemical composition is listed in Table 1. The material exhibited a homogeneous ferritic structure following efficient recrystallization across the entire sample cross-section. Carbide distribution was typically oriented in the sheet forming direction. Inclusions, largely of titanium nitrides and complex oxides, were negligible.

Table 1
Chemical composition of the investigated steel [wt. %]

<table>
<thead>
<tr>
<th>C</th>
<th>Mn</th>
<th>Si</th>
<th>P</th>
<th>S</th>
<th>Cr</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.06</td>
<td>0.50</td>
<td>0.33</td>
<td>0.024</td>
<td>0.002</td>
<td>16.55</td>
</tr>
<tr>
<td>Ni</td>
<td>Cu</td>
<td>Mo</td>
<td>V</td>
<td>Al</td>
<td>Fe</td>
</tr>
<tr>
<td>0.18</td>
<td>0.06</td>
<td>0.02</td>
<td>0.08</td>
<td>0.007</td>
<td>balance</td>
</tr>
</tbody>
</table>

Fig. 1. Structure of nonaffected zone in steel 1.4016

The structural state of the HAZ sub-layers was consistent with the reach and stability of the phase changes in dependence on the height of heating and existing cooling rate.

At temperatures below 850-900 °C, the structure of this type of steel consists of ferrite (both delta and alpha ferrite) and carbides. A fraction of the heat affected region where heating above those temperatures took place falls in the zone of the two-phase ferritic-austenitic structure. At this material heating stage, the austenite evidently formed relatively
fine and uniformly distributed islets and, in comparison with the ferritic volume, was probably partly depleted in chromium. During accelerated cooling within the reach of re-austenitization during welding, this austenitic component transformed into martensite. Thus the final structure of the zone immediately neighbouring the fusion zone consisted of martensite, ferrite and carbides (Fig. 2). The phase composition of this zone was not different from that of the weld metal. There was a difference in the morphology of the grains, which were surrounded by the contrasting martensitic phase, the weld metal containing martensite formed along the boundaries of the columnar structure.

The appreciably rough grain structure along the melting line was due to the intensive growth of the ferritic grain and was formed at temperatures above 1250 °C. Although the subsequent cooling resulted partly in the formation of austenite, the austenitic phase fraction forming from delta-ferrite can be subsequently reduced by reverse transformation to alpha-ferrite. Hence, this phase transformation, in dependence on the particular cooling rates, does not modify the grain boundaries to the full reach, usual in low-alloy steels. The consequences of this process were also observed in the welded joint examined. During the fast cooling across the two-phase region, austenite was formed preferentially at the boundaries of the coarsened grains or, in reaction to the previous intensive grain coarsening, in the direction towards the interior of grains with a certain crystallographic orientation. Complete suppression of the formation of this unstable austenite is not feasible even at high cooling rates. Only very slow cooling following the welding procedure or isothermal hold-up in the phase transformation region (at approximately 900 °C) results in the formation of the ferrite-carbide structure.

In the welded joint tested, natural cooling to room temperature after welding resulted in the transformation of austenite to martensite, whereby its distribution at high temperatures was sustained (Fig. 3).

Carbon and nitrogen based precipitates were also identified. They were very fine precipitates, partly surrounding the sub-grain boundaries. The precipitates visualized locally the consequences of stress relaxation near the melting line. A typical carbide-free zone was formed near the martensitic phase surrounding the coarse grain boundaries.

3.2 Changes in the fracture behaviour

Transgranular ductile fracture only occurred in the welding-unaffected zone of the ferritic steel. The fracture plane morphology was affected appreciably by the presence of carbides. In this manner, the typical "flake"
effect on the fracture plane was observed in the areas rich in carbides (Fig. 4).

Static loading experiments revealed two tendencies in the defective fracture behaviour of the ferritic steel.

Transgranular cleavage fracture occurred locally in the high-temperature segment of the HAZ (Fig. 5). This was the zone where the above transformation of the homogeneous ferritic structure to the two-phase ferritic-martensitic structure had taken place. In contrast to the tendency of carbon diffusion to the grain boundaries in relation to the intergranular steel sensitization, precipitation of carbides near the grain boundaries that were surrounded by martensite was suppressed. On the contrary, fine plate-formed carbides formed in the interior of the coarse ferritic grains. Carbides were thus preferentially formed on the glide planes or at the sub-grain boundaries.

Martensite distribution within the grain coarsening zone appeared to affect the material's fracture behaviour as well. An adverse effect of the occurrence of martensite, consisting in induction of an increased sensitivity to the presence of hydrogen, is generally reported [10]. The experiments revealed a direct effect on the fracture mechanism. In contrast to the predominating cleavage mechanism for the ferritic phase, transgranular ductile fracture mechanism was identified for the microvolumes of the martensitic phase (Fig. 6). The ductile fracture microplanes precisely copied the presence of the low-carbon martensite, i.e. the regions not only at the boundaries but also in the interior of the coarse grains where the austenitic phase had created formations similar to Widmanstatten's structure.

A different defective fracture type was observed in the low-temperature zone of the HAZ. Intergranular fracture took place in the...
region at the transition to the heat-unaffected steel; in the direction towards the unaffected material the intergranular decohesion was combined with transgranular ductile fracture in the zones of presence of the primary carbides (Fig. 7).

Distribution of very fine precipitates was observed at the defective fracture grain boundaries. Analyses of the bare grain boundaries suggested a local higher occurrence of vanadium and silicon. In this context, the coaction of nitrogen, which can react with Fe, Si and Cr in the solid solution, is conceivable. The formation of chromium nitrides – preferentially before iron nitrides – is most likely from the thermodynamic aspect. Precipitation of chromium nitride Cr$_2$N at the grain boundaries following heating above 400 °C is typical of steels of this type [11]. The local hardness differences were measured across a fusion zone of the convenient and inconvenient weld joints; the results are presented as a moving average (Fig.8). Beside the considerable data dispersion in the weld metal there was found out a hardening along the fusion zones. We can assume the dispersion strengthening overcame the coarsed grain influence in both cases. A considerably lower hardness rise was measured at the border HAZ, i.e. in the area of the intergranular fracture. The differences in the width of the HAZ revealed the influence of the thermal exposure.
4. Conclusions

The experiments revealed a tendency towards defective fracture behaviour of non-stabilized ferritic steel associated with the heat-induced changes following welding. The differences between the carbide distributions in the various HAZ sub-layers had a decisive effect on the material's fracture behaviour. Different secondary phase production took place in dependence on the heating intensity and cooling rate. The degradation mechanism in the high-temperature zone is dependent on the final carbide phase distribution.

In the case of a preferential distribution within the grain interior, the relaxation processes along the welding line are limited and the transgranular cleavage fracture mechanism plays a role. The results show that martensite formation in the HAZ of non-stabilized steels has no direct degradation effect on the fracture behaviour of ferritic steel. No cleavage along the ferrite/martensite phase interface was identified; the martensitic phase at the grain boundaries as well as in the grain interior had a ductile fracture mechanism.

Intergranular distribution of secondary phases took place in the low-temperature region at the heat-affected/heat-unaffected interface due to a lower cooling rate. This brought about the observed tendency towards intergranular decohesion. In both cases, defective fracture behaviour, associated with a critical decrease in the weldment ductility, was identified as the limiting degradation mechanism for the steel examined. Based on the relations identified, those processes can be looked upon as an opposite extreme from the aspect of the effect of the diffusion redistribution of carbon, as compared to the processes resulting in intergranular corrosion.

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