ELECTRON BEAM WELDS OF AUSTENITIC STAINLESS STEELS AND ODS STEELS

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Abstract

Ensuring the weldability of advanced heat resistant ODS metallic materials in combination with conventional materials is a prior requirement for their wider use in energy production. The microstructure of ODS steels is composed of alpha iron based matrix with dispersed oxide particles. Due to heating during conventional welding, the microstructure and properties of the resulting weld joints are affected and the joints often become the weakest point of the structure. The electron beam welding with its reduced heat affected zone size may be an answer in this. The presented article is focused on the metallographic evaluation of the structure of heterogeneous electron beam welds which combine austenitic stainless steels with the MA956 ODS steel. EB welded joints were evaluated by light and analytical electron microscopy in the as-welded state and after post-weld heat treatment. Mechanical properties of the weld were evaluated from the results of microhardness profiles. Achieving an appropriate structure of such welds and correct welding parameters are crucial aspects for future successful application of similar joints in energy industry.

Keywords: electron beam welds, ODS steel, oxide particles, heterogeneous joints

1. INTRODUCTION

Increase in efficiency of power generation and chemical engineering is associated with increasing of operating temperatures and the introduction of new technologies (such as thermonuclear fusion). These requirements lead to the development of advanced materials which allow for the design of power plants or equipments whose components are exposed to high temperatures and often work under high stress in corrosive environments [1]. The solution to this problem can be the oxide dispersion strengthened (ODS) alloys. The main potential of these alloys is the increase in the operating temperatures of the device by 100°C. Thermally stable oxide particles dispersed in the matrix are more effective traps for moving dislocations than grain boundaries, thus providing superior high temperature properties compared to conventional alloys [2]. ODS alloys are fabricated by using mechanical alloying of metal (ferritic, ferritic-martensitic and austenitic) matrix with oxide nanoparticles (usually Y₂O₃ or TiO₂). Mechanical alloying is followed by hot isostatic pressing or hot extrusion [1,2,3].

One of the ODS alloys is the ferritic steel INCOLOY MA956. Higher concentration of chromium (above 16%) and concentration of aluminum (about 5%) provide a good corrosion resistance and together with a dispersion of Y₂O₃ particles increase the creep resistance [3,4]. The use of ODS alloys is planned in the next-generation nuclear plants in combination with conventional steels (e.g. 316 austenitic steel) [6].

The choice of suitable welding methods has major influence on degradation of the properties and structural changes of resulting welds. One of the methods with minimal influence on the basic metal is the electron beam (EB) welding. In contrast, with decreasing width of the weld, an increase in the cooling rate occurs and also danger of cracks and pores is higher. For these reasons, setting of welding parameters is crucial for production of defect-free welds [7,8]. Subsequent post-weld heat treatment (PWHT) is performed to reduce the internal stress and to remove undesirable structures after rapid cooling. Correctly chosen conditions of PWHT are also crucial for the resulting properties and overall service life of dissimilar joints [9,10].

In this paper, dissimilar electron beam welds between ODS alloy MA956 and 316Ti steel were prepared. Resulting microstructures were analyzed to evaluate the weldability of these steels by electron beam
welding. The obtained results will be used for further optimization of electron beam welding parameters and for determination of suitable PWHT conditions.

2. EXPERIMENTAL

Welds between ferritic chromium INCOLOY MA956 steel and 316Ti austenitic stainless steel were used in the present study. The chemical compositions of the used steels are given in Table 1. Cylindrical samples with diameter of 12 mm were welded by electron beam without any use of filler material. The welding direction is perpendicular to rolling direction of the base metals. Welds were processed by one pass of EB at 55 kV accelerating voltage, 15 mA beam current and 10 mm/s welding speed by using FOCUS MEBW-60/2 EB welding facility. MA956/316Ti welds were evaluated in as-welded state (AW) and in state after post-weld heat treatment (PWHT). PWHT consisted of one tempering treatment at 750 °C in argon atmosphere, followed by slow cooling in Ar atmosphere.

Structure of base metal (BM), weld metal (WM) and the heat-affected zone (HAZ) were evaluated using light microscopy (LM) and scanning electron microscopy (SEM). SEM was also used to analyze the chemical composition of the EB welds. From the measured concentrations EDS maps of alloying elements were prepared, also the alloying elements average concentration was determined and compiled in sequences across the welded joints. In addition, microhardness measurements were performed for comparison with the results obtained from light and electron microscopy. Microhardness measurements HV0,1 was performed on semiautomatic hardness tester LECO LM 247 AT in perpendicular direction to the axis of the weld. The LM Zeiss Axiovert Z1 and Zeiss UltraPlus SEM were used in the analyses.

Table 1 Chemical composition of base materials (Wt. %)

<table>
<thead>
<tr>
<th>Elements</th>
<th>C</th>
<th>Cr</th>
<th>Ni</th>
<th>Al</th>
<th>Mo</th>
<th>Mn</th>
<th>Si</th>
<th>Y2O3</th>
<th>Ti</th>
<th>V</th>
<th>Nb</th>
<th>N</th>
</tr>
</thead>
<tbody>
<tr>
<td>MA956</td>
<td>0,016</td>
<td>20,00</td>
<td>-</td>
<td>4,50</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>0,50</td>
<td>0,50</td>
<td>-</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>316Ti</td>
<td>0,02</td>
<td>17,10</td>
<td>11,80</td>
<td>-</td>
<td>2,25</td>
<td>1,83</td>
<td>0,60</td>
<td>-</td>
<td>0,19</td>
<td>0,14</td>
<td>0,02</td>
<td>0,06</td>
</tr>
</tbody>
</table>

3. RESULTS AND DISCUSSION

Electron beam (EB) welds exhibit classical nail-like appearance [7]. Maximum width in head of the weld is in the range 1,3 to 1,5 mm. Significant HAZ was not observed in both base materials. Electron beam welds can be classified into five different areas: base metals, weld metal and transition region (TR) on each side of the base metals. EB welds generally contain a large amount of defects. Across the weld there were observed pores of different sizes. Cracks occurred in WM and along the boundaries of ferritic grains close to TR on MA956 BM side. Cracks in the base material formed during cooling the WM due to tensile stresses. Presence of defects in WM is associated with improper setting of the EB welding parameters. This parameters setting, in case of materials with different physical properties, is very difficult and requires a numerous test welds to optimize.

When evaluating BM of chromium steel INCOLOY MA956 fully ferritic structure was observed (Fig. 1a). Due to the extrusion elongated grains were grown with size of about 100 µm in perpendicular direction to extrusion. The grain size ratio is about 100:1 (parallel: perpendicular to direction of extrusion) in accordance to [1]. Presence of pores on the grain boundaries was observed. Fig. 1b shows microstructure of 316Ti steel which is formed by equiaxial austenitic grains with average size around 50 µm. In most of the austenitic grains twins were observed, likely from previous forming and heat treatment of BM. In the microstructure irregularly shaped nitrides and carbonitrides of titanium were observed. The latter is due to the addition of Ti as stabilizer in the stainless steel. The used PWHT hadn’t significant influence on the structure of both BM.

Comparing Fig. 2a and 2b allows determining the effect of PWHT on changes in the structure of WM and TR on 316Ti side. AW sample has a finer structure than the sample after PWHT. Size the grain in AW sample is determined by the cooling rate of WM. Coarsening of the equiaxial grains in WM was observed after PWHT.
These grains should be stable at higher temperatures due to presence of the oxide particles. The electron beam welding leads to dissolution and subsequent coarsening of dispersion strengthening particles [5,9].

Fig. 1 Structure of MA956 base metal (a), structure of 316Ti base metal (b).

Fig. 2 Structures of EB welds AW (a) and PWHT sample (b).

Fig. 3 MA 956/weld metal interface with crack (a). WM/316Ti interface of AW sample (b). WM/316Ti interface of PWHT sample (c) with detail (d).

HAZ wasn’t observed on interface between ODS steel and WM in both samples. BM directly passes through the fusion zone in WM. During crystallization epitaxial growth of WM grains at this interface was observed
(Fig. 3a). Interface between 316Ti BM and WM (Fig. 3b) shows a narrow HAZ (approximately 0.2 mm), which is characterized by structure with coarser austenitic grains and secondary phases particles along grain boundaries near the interface with WM. The lowest density of twins in austenitic grains was observed in this HAZ. The most significant changes were observed after PWHT in this area. The HAZ is wider and contains minimum amount of austenitic twins. At the same time the amount and shape of secondary phases at WM/316Ti interface was changed (Fig. 3c and 3d).

To compare the results obtained from light and electron microscopy microhardness measurements were performed across weld joints (Fig. 4). When comparing the microhardness of the EB weld, measured values could be linked with the observed structures. Significant changes weren’t observed in the ODS steel BM, microhardness range is between 263 and 295 HV0.1. Slight decrease in microhardness values was observed at ODS/WM interface. This can be explained by melting and coarsening of oxide particles in the vicinity of interface during EB welding. Average microhardness value of WM in AW sample was 497 ± 11.5 HV0.1. After PWHT, significant decreases in average microhardness of WM was observed at 370 ± 8.8 HV0.1. These changes in microhardness correspond with the observed structures of AW and PWHT samples (Fig. 2a and 2b). The presence of secondary phase did not significantly influence the values of microhardness across WM/316Ti interface. However, due to annealing of the austenitic structure in the vicinity of WM during EB welding the decrease of microhardness was observed to 178 – 179 HV0.1 in TR on side of 316Ti. Average microhardness values of austenitic BM of each sample was 209 ± 7.1 HV0.1 (AW) and 215 ± 14.0 HV0.1 (PWHT).

EDS analysis of the main elements (Cr, Ni, Mo, Ti, Al) in EB welds was carried in same direction as the microhardness measurements. Comparing the chemical composition of both welds the dilution of both BM’s was demonstrated (Fig. 5 and 6). Individual parts of welded joints can be distinguished by the measured concentrations of individual elements through these regions. Significant changes weren’t observed in the concentrations of alloying elements in AW and PWHT samples. In chromium concentration sharp transition at ODS/BM interface can be observed. This step is reduced after PWHT. Step change in content of nickel is also observed at both interfaces of the WM with BM’s. Changes of chemical composition profiles can be seen after PWHT. These changes are caused by diffusion of alloying elements in the direction of concentration gradient. Common feature of these interfaces is the expansion of areas with concentration changes. At WM/316Ti interfaces significant step on profile of Ni concentration was observed. This step corresponds with the observed structure at this interface in PWHT sample (Fig. 3d). The behavior of other elements is similar - redistribution in direction of gradient of concentration. Decrease of intensity of the step changes of concentrations and the expansion of transition areas occurred after PWHT.

![Fig. 4 Microhardness measurements across MA956/316Ti EB welds.](image)

Fig. 7 shows EDS maps of the distribution of alloying elements on WM/316Ti interface in AW sample. Redistribution of Cr and Mo in to secondary phase along austenitic grain boundaries may cause precipitation of undesirable phases (σ-phase and Laves phase). Further it shows the presence of Ti and Al particles in WM. Concentrations of alloying elements were partially equalized in each phase at WM/316Ti interface after PWHT (Fig. 8). However, during long-term exposition and due to different chemical composition of welded
steels, the redistribution of interstitial elements may occur. Further evaluation and modeling of diffusion processes of interstitial elements is recommended in accordance to [11].

**Fig. 5** Concentrations of alloying elements across MA956/316Ti weld (AW).

**Fig. 6** Concentrations of alloying element across MA956/316Ti weld (PWHT).

**Fig. 7** EDS maps of WM /316Ti interface (AW).

**Fig. 8** EDS maps of WM /316Ti interface (PWHT).
4. CONCLUSIONS

This study shows the possibility of welding the ODS MA956 alloy and 316Ti alloy by EB welding and the requirement of subsequent PWHT. The main limiting factor for wide use of this welding technique is the knowledge of appropriate welding parameters to ensure defect-free welds and also the dissimilar physical properties of each alloy. During metallographic evaluation of both samples no presence of undesirable structures was observed. The measured values of microhardness correspond to observed microstructures. Used PWHT hadn’t significantly impacted on concentration of alloying elements in different parts of EB joints, only reduced the concentration step at each interface. To optimize the structure and properties of dissimilar EB weld joints MA956/316Ti further evaluation of PWHT conditions and their influence on resulting welds is needed. However, the first prerequisite for successful welding of dissimilar base metals is finding of suitable welding parameters. Further evaluation of long-term exposure at evaluated temperatures for these joints is necessary before industrial applications can be considered.

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