In stainless-cladded carbon steels, microstructural changes, due to carbon and other elements diffusion, take place at the interface strictly depending on the temperature values reached during the production process. The microstructural changes at the interface in a carbon steel (ASTMA 515 Gr 60) cladded by hot rolling with an AISI 304L steel was examined in this paper. Carbide precipitation, due to carbon diffusion towards the austenitic steel, was observed by optical and scanning electron microscopy (SEM); substitutional elements distribution was studied by energy dispersive spectroscopy (EDS); mechanical properties at the bimetallic interface were tested by Vickers microhardness measurements. These measurements showed a hardened region at the stainless side near the interface: here, during production process, Cr carbide precipitation occurs at the austenitic grain boundaries, with a partial local Cr depletion.

**Abstract**

In stainless-cladded carbon steels, microstructural changes, due to carbon and other elements diffusion, take place at the interface strictly depending on the temperature values reached during the production process. The microstructural changes at the interface in a carbon steel (ASTMA 515 Gr 60) cladded by hot rolling with an AISI 304L steel was examined in this paper. Carbide precipitation, due to carbon diffusion towards the austenitic steel, was observed by optical and scanning electron microscopy (SEM); substitutional elements distribution was studied by energy dispersive spectroscopy (EDS); mechanical properties at the bimetallic interface were tested by Vickers microhardness measurements. These measurements showed a hardened region at the stainless side near the interface: here, during production process, Cr carbide precipitation occurs at the austenitic grain boundaries, with a partial local Cr depletion.

**INTRODUCTION**

Cladding is an economical alternative to expensive high-alloy solid plate, because it combines the base material mechanical properties with wear and/or corrosion resistance of the surface materials. Surface layers of nickel alloy or stainless steel on a carbon steel substrate can be produced by fusion welding [1] or solid state welding, such as hot mechanical pressing [2] and rolling [3]. During the production process, due to wide differences in composition between base steel and cladding materials, considerable transport of elements may occur at the bimetallic interface, which could result in precipitation of phases. Computer simulations were developed to predict diffusion and microstructural changes close to the interface [4]; but the thermodynamic data utilized for numerical applications cannot have a general validity, because are strictly depending on temperature and other process parameters.

In particular, microstructural changes at the interface, due to carbon and other elements diffusion, are affected by the temperatures reached during the production process. Microstructural analysis of the interfaces obtained by solid state welding of different combinations of low alloy steels to nickel alloys, using various techniques (hot uniaxial pressing, hot plane strain compression, hot isostatic pressing and coextrusion), was carried out in [5,6,7]. In [8] carbide precipitation was observed at the interface between carbon and austenitic steel, cladded by hot rolling.

In this work the microstructural changes at the interface of a carbon steel cladded by hot rolling with an austenitic stainless steel was examined by optical and SEM observations, with EDS microanalysis, to study the diffusive phenomena concerning with the substitutional elements; the mechanical properties near the bimetallic interface were tested by Vickers microhardness measurements. In particular the nature of the interface between the two steels, the carbon and substitutional elements diffusion and the carbide distribution at the austenitic grain boundary were investigated and discussed.
The material examined was a C-steel plate (6.5 mm thick), cladded by hot rolling with an austenitic stainless steel (2.5 mm thick), according to a patented process [3]. The base material specification is ASTM A 515 Gr. 60; the cladding material specification is AISI 304L. The respective compositions are given in table 1.

Metallographic observations were carried out on samples cut from plates in as supplied conditions. The metallographic samples were prepared by conventional grinding and polishing methods and then etched by Nital (2% HNO₃, 98% ethyl alcohol) and by Glyceregia (42% glycerol, 42% HCl, 16% HNO₃) to reveal respectively ferritic and austenitic grains.

The microstructure features of the two steels near the cladding line were characterized by optical and SEM observations. The scanning electron microscope was of type JSM-5600LV, equipped with an energy dispersive spectrometer. Two accelerating voltage values were considered: 25 and 12 kV, the lower one was utilized to reduce the electron beam interaction depth inside material and investigated Cr distribution around carbides.

Vickers microhardness tests (25 g load and 10 s time) were performed along traverses crossing the bimetallic interface to show the changes of the mechanical properties. A number of tests for each distance from the cladding line was performed to have an uncertainty (given by the estimation of standard deviation on measured values, divided by the square root of the number of tests), less than ±5 HV at the ferritic side and ±7 at the austenitic side.

EDS measurements on traverses across the cladding line have shown the diffusion profiles of Cr, Ni, and Mn (fig. 3). These profiles are centered around the boundary between ferritic and austenitic grains, that are observable inside the cladding line (fig. 2b); the diffusion distances of the substitutional atoms at the ferritic side can be evaluated as about 20 µm; so only the ferritic grains adjacent to the cladding line are affected by these atoms.

**Table 1. Base steel and clad steel compositions (*)**

<table>
<thead>
<tr>
<th>Composition (% in weight)</th>
<th>C</th>
<th>Mn</th>
<th>Si</th>
<th>P</th>
<th>S</th>
<th>Al</th>
<th>Cr</th>
<th>Ni</th>
</tr>
</thead>
<tbody>
<tr>
<td>Base steel:</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>ASTM 515 Gr 60</td>
<td>0.14</td>
<td>0.85</td>
<td>0.20</td>
<td>0.008</td>
<td>0.001</td>
<td>0.04</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>Clad steel:</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>AISI 304 L</td>
<td>0.017</td>
<td>1.32</td>
<td>0.39</td>
<td>0.029</td>
<td>0.003</td>
<td>-</td>
<td>18.39</td>
<td>10.07</td>
</tr>
</tbody>
</table>

(*) Inspection certificate by Voest-Alpine in Linz (Austria).

**RESULTS**

SEM observations, performed near the bimetallic interface, show microstructural modifications in the ferritic-perlitic base material and in the austenitic cladding material. A band, clearly detectable, but with a very little width (less than 10 µm), divides the two steels and can be named as “cladding line”. As can be observed after Nital etching (fig. 1), the cladding line follows the profile of the ferritic grains. Near the bimetallic interface, the carbon steel exhibits a decarburized zone, about 120 µm wide, without perlitic grains and with large ferritic grains having an average size of 30-50 µm, about twice the ferritic grain size far from the cladding line. Far from the cladding line, the perlitic grains are arranged along rolling band parallel to the bimetallic interface.

SEM observations on the austenitic stainless steel side, after Glyceregia etching, show a carburized region (about 200 µm wide) with carbide precipitation at the grain boundary (fig. 2a-b).

In particular the carbide concentration gradually decreases with increasing the distance from the cladding line. Near the bimetallic interface the austenitic grains appear deformed, with larger size (about 30 µm) parallel to the cladding line. At a distance greater of about 100 µm from the cladding line, the austenitic grains are undeformed with an average size of about 20 µm, indicating that recrystallization occurred.

Fig. 1: SEM micrograph of the bimetallic interface, C-steel side after Nital etching

Fig. 2: SEM micrograph of the bimetallic interface, austenitic side after Glyceregia etching: a) general view; b) detail of the cladding line

Fig. 3: EDS spectrum of the cladding line.
Along a traverse across the bimetallic interface, starting from the C-steel base material, the following regions can be observed:

1. The base steel, with microstructure, far from the bimetallic interface, with ferritic and perlitic grains of short size; in particular with the perlitic grains located along rolling bands. This microstructure is depending on rolling parameters, heat treatment and cooling rates.

2. A decarburized region in the base steel, close to the interface, with large ferritic grains. This is a consequence of a carbon copious diffusion towards the austenitic steel.

3. A narrow band (cladding line), parallel to the original interface, that follows the ferritic grain profiles and separates the base steel from the austenitic layer. Inside the cladding line it is possible to distinguish the interface between the large ferritic grains and the deformed austenitic grains. The substitutional atoms diffusion profiles are centered across this interface. Because the Ni diffusion profile is very sharp, a shift of the austenitic structure towards the C-steel is not observed. This shift to the cladding line were interested by the substitutional elements diffusion. The substitutional atoms diffusion towards the ferritic side is compensated by a contrary flow of Fe atoms.

Although the carbon concentration profile has not been determined by EDS measurements, the width of the region of carbide precipitation at the austenitic side indicates that carbon diffuses over larger distances than substitutional elements. EDS measurements, performed with an accelerating voltage of 12 kV to reduce the thickness of material involved in the X-ray emission, have given the Cr distribution around carbide precipitates characterized by high Cr content (fig. 4): while in the bulk of the austenitic grain the Cr content is quite constant, a small zone of Cr depletion, with a relative minimum of Cr content, can be observed at grain boundaries near carbides. The Vickers microhardness tests, performed on a traverse along the bimetallic interface (fig. 5), give a remarkable peak value in the austenitic side near the cladding line of about 350 HV (fig. 6). On the austenitic side, after this peak, the hardness profile gradually decreases down to values of about 250 HV; while, on the ferritic side, it suddenly decreases down to a minimum value of about 160 HV, located very close to the cladding line, and then increases to the typical value of the base C-steel of about 200 HV.

**DISCUSSION**

Along a traverse across the bimetallic interface, starting from the C-steel base material, the following regions can be observed:

1. The base steel, with microstructure, far from the bimetallic interface, with ferritic and perlitic grains of short size; in particular with the perlitic grains located along rolling bands. This microstructure is depending on rolling parameters, heat treatment and cooling rates.

2. A decarburized region in the base steel, close to the interface, with large ferritic grains. This is a consequence of a carbon copious diffusion towards the austenitic steel.

3. A narrow band (cladding line), parallel to the original interface, that follows the ferritic grain profiles and separates the base steel from the austenitic layer. Inside the cladding line it is possible to distinguish the interface between the large ferritic grains and the deformed austenitic grains. The substitutional atoms diffusion profiles are centered across this interface. Because the Ni diffusion profile is very sharp, a shift of the austenitic structure towards the C-steel is not observed. This shift...
was observed in [5,6,7] for different combinations of low alloyed C-steel cladded with Ni alloys having higher Ni contents than the austenitic steel here considered. However the measured atoms diffusion and the width of the carbide precipitation region, due to a large carbon diffusion, indicate in all instances that metallurgical bonding gives interdiffusion of elements.

In particular it is interesting to observe that, beyond the interface, the carbide precipitation begins immediately at the boundary of the first austenitic grains. Considering that carbon diffusion in the austenitic steel leads to carbide precipitation as soon as bonding begins, it is reasonable to assume that the bimetallic interface observed inside the cladding line coincides with the original interface between the two steels.

4. A zone rich of carbide precipitates in the stainless steel near the cladding line. Due to the high Ni and Cr contents, the solubility of carbon is low and a copious Cr carbide precipitation takes place on twins and grains boundaries, where the pipe diffusion is more rapid if compared to the bulk diffusion. The width of the zone of carbide precipitation depends on bonding temperature and on time of cooling, according to the process parameters.

In the zone of carbide precipitation, a typical microstructure of rolled austenite indicates that recrystallization after hot working has not occurred. Moreover Cr depletion at grain boundaries, near carbide precipitates, took place making the austenite sensitized to an intergranular corrosive attack.

5. The stainless steel region far from the cladding line with microstructure characterized by recrystallized equiaxed austenitic grains and free from carbide precipitation.

The experimental profile of each diffused element across the bimetallic interface can be fitted by an error type function, as follows:

\[ c(x) = \left( \frac{c_1 - c_2}{2} \right) \text{erf} \left( \frac{x}{2\sqrt{D} t} \right) \]

where \( c_1 \) and \( c_2 \) are the starting concentrations of the considered element in both materials, respectively, \( x \) is the distance from the bimetallic interface, \( t \) is the bonding time and \( D \) is the diffusion coefficient.

This expression characterizes the element transport controlled by solid state diffusion. Making the assumption that the diffusion coefficient does not change with concentration, from the \( D t \) products which give the best fit of the experimental profiles, the \( D \) coefficients of Cr, Ni and Mn can be obtained. The diffusion coefficients, calculated for a processing time of 1 h (table 2), agree to data published in [7] for \( T = 1100^\circ \text{C} \). This value is consistent with the temperatures reached during the production process. Even if carbon diffusion can not be detected by microanalysis measurements, its effects at the austenitic side are showed by microhardness test. The microhardness profile agrees with the observed morphology: the experimented hardness peak is in the carburized region at the austenitic side, while a minimum of hardness is recorded in the decarburized region at the C-steel side, characterized by large ferritic grains: here the hardness decreases, even if some migration of alloy elements (Ni, Cr, Mn) occurs.

CONCLUSIONS

Metallurgical bonding at the interface between C-steel base material and stainless steel cladding material, obtained by hot rolling, gives rise to the interdiffusion of carbon towards the austenitic side and substitutional elements towards the ferritic side. Microstructural changes near the bimetallic interface have been observed: a decarburized region at the C-steel side; a hardened region with high C content at the stainless steel side; here Cr carbide precipitation occurs at the austenitic grain boundaries which consequently undergoes a partial Cr depletion.

ACKNOWLEDGEMENTS

Our thanks are due to the management of the Voest-Alpine Stahl Linz Industrie for the cladded material supplied.

REFERENCES


Table 2. Diffusion coefficients

<table>
<thead>
<tr>
<th>Element</th>
<th>Ni</th>
<th>Cr</th>
<th>Mn</th>
</tr>
</thead>
<tbody>
<tr>
<td>D (m/s)</td>
<td>1.2E-11</td>
<td>3.3E-11</td>
<td>5.5E-11</td>
</tr>
</tbody>
</table>