Interface layer deformation thickening analysis of hot rolling carbon steel–stainless steel clad plate

Haibin Li, Hongmei Zhang, Jianmei Wang, Meirong Shuai, Xiaogang Wang, Yugui Li and Li-Feng Ma

School of Material Science and Engineering, Heavy Machinery Engineering Research Center of Education Ministry, Taiyuan University of Science and Technology, Taiyuan 030024, Shanxi, People’s Republic of China
2 Steel Rolling Equipment Subco., Taiyuan Heavy Machinery Group Co., Taiyuan 030024, Shanxi, People’s Republic of China
3 Author to whom any correspondence should be addressed.
E-mail: 285289934@qq.com

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Abstract

In this study, a stainless steel – carbon steel clad plate was rolled on a two-high mill. The microstructure and thickness of the bonding interface before and after the tensile test were observed through scanning electron microscopy. Results showed that the sample underwent plastic deformation via slipping in the static tension test, resulting in the rapid decrease in the Schmid factor (SF). However, the weighted average value of 1/SF increased from 2.1 to 2.2 after the tensile test. Consequently, the size of the grains with diameters between 12 and 120 μm decreased, whereas the percentage of grains with diameters of less than 5 μm sharply increased to 76.7%. The carbide layer in the bonding interface easily produced stress concentration and formed a fibrous region during the subsequent tensile test. The carbide particles in the bonding interface extended along the slip direction rather than the tensile direction, resulting in increased carbide thickness. The carbides hardly entered the SS but easily extended into the soft ferrite area of the CS surface. The deformation energy increased with a decreased in grain size, and the increase in the part size was the increment of the surface energy of the small grains.

1. Introduction

Stainless steel (SS)/ carbon steel (CS) clad plates are bonded in a specific manner; the former serves as the clad layer whilst the latter serves as the substrate. This type of clad plate possesses numerous characteristics, including aesthetic property, corrosion resistance, good mechanical properties of SS and low cost and high strength of CS [1, 2]. These clad plates are in high demand and are widely used in petroleum and chemical industries [3]. At present, clad plates are mainly produced through the explosion method. However, these plates cannot be manufactured in a large scale because of the restrictions of production technology, including severe pollution, high cost and the limited size of clad plates [4, 5]. Moreover, these clad plates have received considerable attention in the context of product quality in terms of the shrinkage cavity, crack and stomata of the bonding interface under explosive blast waves [6, 7]. Similarly, rolling technology has attracted wide attention. This technology presents the advantages of economical and efficient production capacity, low pollution and low energy consumption [8, 9]. However, the binding rate, bond quality and heat treatment process exhibit certain deficiencies under production technology [10–12]. For instance, the interface is easily oxidised at a high temperature, thereby reducing the strength of the composite interface.

Vacuum hot rolling bonding (VHRB) has been applied to the production of clad plates to effectively prevent oxidation [13]. Previous studies mostly focused on bonding interface microstructures and mechanical properties, but the microstructure evolution before and after tensile tests is seldom reported. The base and flyer materials utilised in the present study are CS (Q235A) and austenitic SS (304), which are arranged and welded through VHRB technology. The CS and SS plates’ thicknesses are 5.5 and 2.5 mm, respectively. CS mainly
consists of C (0.1%), Si (0.3%), Mn (0.5%), S (0.03%) and Fe (Bal.), whereas SS comprises C (0.038%), Si (0.43%), Mn (1.21%), S (0.002%), P (0.022%), Cr (18.23%), Ni (8.03%) and Fe (Bal.). The bonding interface microstructures before and after the tensile test were observed through scanning electron microscopy (SEM), and the change mechanism of the bonding interface under tensile stress was analysed.

2. Experimental method

The specifications of the three- and four-ply bond billets were 140 mm × 115 mm × 10 mm and 150 mm × 125 mm × 15 mm, respectively. The three-ply specimen (SCS) was composed of CS in the middle and SS in the outer surface; no release agent was used between layers. Meanwhile, the four-ply specimen (CSSC) was composed of CS in the upper and lower layers and SS in the middle two layers. Between two layers, SS was coated with a release agent to form a nonbonding surface that should separate the two clad plates smoothly after the symmetrical rolling process.

All experiments were carried out on a two-high mill with a diameter of 320 mm and a length of 350 mm. The rolling velocity was 0.1 m s⁻¹, and the press quantity was 30%. The billets were heated to 1,300 °C in the heating furnace and soaked for 15 min. The rollers were maintained at normal atmospheric temperature. After the deformation, the clad plates were cooled to room temperature in air. The Morphologies of the three-ply and four-layer plates are shown in figure 1. After the welded seam was sawed off from the clad plate and a small billet was cut from each plate along the rolling direction, the billet was made into metallographic specimens. Then, the specimens were etched separately in a solution containing a mixture of 4 ml of HNO₃ and 96 ml of C₂H₅OH. The microstructures were detected using a scanning electron microscope (model S4800 type) equipped with an energy dispersive spectroscopic (EDS). The tensile tests were conducted using a WDW-100 electronic universal tensile testing machine.

3. Results and discussion

3.1. Microstructure of bonding interface layer

The bonding interface micrographs of the three- and four-ply plates are shown in figures 2(a) and (b), respectively. In figure 2(a), only ferrite precipitated in the CS near the bonding interface, as observed in previous research [14]. In addition, abundant fine carbide particles precipitated in ferrite. The flake was formed in an orderly manner in the CS/SS bonding interface, with an approximate thickness (δ) of 5 μm. Within the SS plate, mass holes with different kinds, sizes and shapes were observed possibly due to the atomic diffusion of the Kirkendall effect.

Figure 2(b) shows that the ferrite grain was flat and free from precipitated carbide particles. However, the micropore in the SS was smaller than that in the three-ply plate. Similarly, the flake layer of the interface was formed, and its δ was approximately 4 μm.

The EDS analysis indicated that the interfacial flake compositions of the three- and four-ply plates were approximately identical and comprised Fe, C and Cr atoms. For instance, the mass fractions of the four-ply specimen (marked by P) comprised C (10.78%), Cr (6.26%) and Fe (82.95%).

After the bonding interface of CS/SS was detached using a special method, the surface of CS was analysed through x-ray diffraction (XRD), which indicated the presence of ferrite and M₇C₃ (figure 3(a)). The ferrite...
served as the CS matrix, and carbide \( \text{M}_7\text{C}_3 \) was the main precipitate on the surface of CS which formed from the Cr atoms caused by the SS diffusion into the surface of CS after hot rolling. Hence, carbide \( \text{M}_7\text{C}_3 \) must be contained in the flake or ferrite of the bonding interface (figure 2). The SS surface of the sample was subsequently analysed, and only austenite (\( \gamma \)) and ferrite were found (figure 3(b)).

3.2. Fracture morphology of bonding interface layer

A tensile test was conducted using tensile strengths of 498.7 and 470.2 MPa, both of which exceeded the minimum tensile strength of the CS/SS clad plate according to the international standard of GBT8165-2008. As shown in figure 4(a), the fracture morphology near the bonding interface of the three-ply specimen involved smooth brittle fracture, which was different from the reticular dimples of the CS matrix morphology. The dense carbide flake easily produced stress concentrations, resulting in fragments which then formed a fibrous region in the bonding interface. By contrast, figure 4(b) consisting of several dimples due to the SS matrix presented good plasticity. The fracture morphology of the four-ply specimen showed that the CS surface on the left figure presented several dimples, which are not an absolute characteristic of brittle fracture. In addition, the bonding interface had a fibrous structure (figure 4(c)). The right fracture surface of SS presented dense dimples.

The specimen broke under a small force increment of \( 1 \text{ N-mm}^{-2}\text{s}^{-1} \) after the yield stage. The thicknesses of the bonding interface layer of the three- and four-ply specimens (figure 4) considerably increased relative to those of the original specimens (figure 2).

3.3. Effect of stress on the thickness of the bonding interface layer

The carbide microstructure and layer thickness of the fracture were altered by the stress concentration during the static tension test; such change could also affect an area without necking under an even stress. A piece with a uniform deformation zone from the fracture of 2–3-cm distance was cut, and the microstructure was analysed. The carbide flake almost turned into fine particles, and only few flakes existed near SS (figures 5(a) and (b)). The boundary of carbide and SS was clearly without changes. Moreover, a part of the unwounded ferrite grain boundaries entered the interior of the carbide layer and became surrounded by fine carbide particles. The
thickness of the carbide layer extended after the static tension test. The average widths of the three- and four-ply specimens were 15 and 12 μm, respectively; these values were three times higher than those of the original specimen. The morphology of a specimen changes under static tension at room temperature, but the eutectoid transformation remains constant [15].

Several studies [16, 17] have been conducted to investigate the grain boundary diffusion induced by heat treatment or electro-beam irradiation. The results showed that the segregation and diffusion of solute atoms at the grain boundaries can increase the interfacial thickness by approximately several nanometres. However, after the tensile test in the current work, the carbide diffusion of the clad plate’s bonding interface increased the layer thickness by several micrometres and thereby exceeded the diffusion result under heat treatment or electro-beam irradiation. The temperature in the static tension test was markedly lower than that in heat treatment or electro-beam irradiation [16, 17]. The atom diffusion at room temperature negligibly inhibited lattice distortion and decreased dislocation density. In addition, the carbide flake of the bonding interface broke down into small particles during the static tension test, and the mechanism of the carbide layer thickening differed from previous mechanisms.

Figure 4. Tensile specimens and surface morphologies of the tensile specimens: (a) Tensile specimens and sampling positions; (b) SCS; (c) CSSC.

Figure 5. Bonding interface microstructure of the tensile specimen: (a) SCS; (b) CSSC.
3.4. Influence of stress on carbide microstructure

The boundary of carbide and SS after the tensile specimen test was clearly without changes, and the small carbide particles were embedded in the ferrite grain (figure 5). Therefore, the CS near the bonding interface was the main research object. The substrate CS was a body-centred cubic solid solution, where the base plane \(\{110\}\) was a slip plane and the slip direction was \([111]\).

When the base plane of CS was subjected to yield stress \(\sigma\), the angle between the stress axis of the base plane and the normal axis of the slip plane \(n\) was \(\theta\), whereas that between the stress axis and the slip direction was \(\beta\) (figure 6). The shear stress \(\tau\) in the direction of the slip is expressed as

\[
\sigma = \frac{\tau}{\cos \theta \cos \beta},
\]

where \(\cos \theta \cos \beta\) is the Schmid factor (SF) which is used to distinguish whether or not the grain of the matrix easily slips.

When \(\tau = 1\),

\[
\sigma = \frac{1}{\cos \theta \cos \beta} = \frac{1}{\text{SF}}.
\]

The experimental data were investigated using the ZEISS Sigma field emission scanning electron microscope based on electron backscattered diffraction (EBSD) with an electron acceleration voltage of 30 kV and a scanning step of 0.2 \(\mu m\). The observation surface of the EBSD was in the rolling–normal direction surface. The SF near the bonding interface of the four-ply specimen before and after the tensile test is displayed in figure 7.

As shown in figure 7(a), the percentages of SF equal to, higher than and lower than 0.475 in the original sample were approximately 50.85%, 12.6% and 37%. This finding indicated that the CS produced new grains during recrystallisation and that such grains eliminated the defects inside the grains caused by slipping during the hot rolling process. The specimen easily deformed under stress during soft crystal orientation because the SF near the bonding interface was high. The SF of the tensile specimens is shown in figure 7(b). The percentage of SF equal to 0.465 yielded the largest value, which was then reduced to 22.8%. The percentage of SF more than 0.465 was 20.7%. However, most SF values were less than 0.465; thus, this percentage increased to 52.5%. The sample experienced plastic deformation via slipping in the static tension test, leading to decreased \(\theta\) or \(\beta\) and SF. This result indicated that the grains of CS slipped along the slip plane and that the specimen after the tensile test hardly deformed relative to the original specimen.
The grain size near the bonding interface of the four-ply specimen was analysed and calculated using the software CHANNEL5 SP9B (figure 8). Figure 8(a) shows that before the tensile test, the percentage of small grains with a diameter of less than 12 μm was 55%, that of medium grains with a diameter between 40 and 120 μm was 25%, and that of large grains with a diameter between 172 and 210 μm was 20%. However, the grain size of the tensile specimens decreased; the percentage of grains with a diameter of less than 5 μm sharply increased to 76.7%, whereas the remaining 23.7% were composed of grains with a diameter of less than 26 μm (figure 8(b)).

A recent study [18] reported that the mechanical energy provided by external load transformed into internal crystal energy, which was stored in grains in the form of strain energy for slipping. Furthermore, the grain storing strain energy produced cracks as the stress increased. These cracks were filled up by the surrounding matrix material, and the deformation-induced plasticity reduced the stress level near the crack tip [19]. In conclusion, all grains slipped and broke down into small grains after the tensile test. The carbide M7C3 consists of large and hard particles [20], which easily shatter into fragments during deformation. Therefore, the carbide M7C3 in the CS must be broken down into small particles during the static tension test. As previously mentioned, the grain cracks were filled up by the surrounding matrix material, and the small particles broken down from the large particles were covered in ferrites.

The yield stress is determined by using the Hall–Petch (H–P) formula, that is,

$$\sigma = \sigma_0 + K\sqrt{d},$$

where $\sigma_0$ is the lattice friction, $K$ is the H–P constant and $d$ is the grain size.

Combining equations (2) and (3) yields

$$\frac{1}{SF} = \sigma_0 + K\sqrt{d}.$$

The value of $1/(SF)$ is proportional to $1/d^{3/2}$, as shown in equation (4); both of them can be calculated through the efficacy coefficient method using the result of EBSD (figures 7 and 8). Then, the $H$–$P$ constant with slope $K$ can be obtained by linear fitting. The $K$ value of the original specimen was 1.12, which increased to 1.11 after the tensile test, indicating that $K$ increased with a decrease in grain size. Furthermore, $\sigma_0$ increased from 1.46 to 1.94 after the tensile test. The maximum value ($1/(SF)$) of the original specimen decreased from 2.6 to 2.4 (figure 9) and became concentrated between 2.0 and 2.4 after the tensile test due to the decrease in grain size. However, the weighted average value of $1/(SF)$ increased from 2.1 to 2.2 after the tensile test, thus indicating that grain refinement enhanced the yield strength.

Large particles easily shatter into fragments during deformation [21]. The flake elongates along the vertical section as stress increases; otherwise, the flake breaks. The flake slips along the stress ($\sigma$) direction after breaking, whereas the loading capacity of the carbide flake improves as the distortion strengthens (figure 6). Therefore, with an increased stress, the flake continually slips in the vertical section until the flake breaks down into small pieces. In this work, the carbide M7C3 particles in CS slipped to both sides of the stress direction, leading to the extension and thickening of the carbide layer along the direction perpendicular to the tensile axis.

Figure 10(a) shows that the phase of SS before the tensile test was austenite (A) and that carbide did not exist in SS. Few carbide M7C3 particles were dispersed in the ferrite (F) of the CS matrix. The phase of the tensile specimen was the same as that of the SS of the original specimen, but a large amount of M7C3 was observed in the CS matrix (figure 10(b)).
The microhardness of the three- and four-ply original specimens was tested using an HDX-1000 type Micro-Vickers sclerometer. The values for the CS ferrite were 162 and 146 HV, whereas those for carbide were 195 and 171 HV. The values for SS were 251 and 231 HV. Figures 5 and 10 suggest that the carbide hardly entered the SS and only extended into the soft ferrite of the CS surface, thereby increasing the thickness of the bonding interface. The maximum thickness was located in the fracture of the sample, where the grains were most likely to break when the stress attains reached the maximum value.

The strip-like grains of the hot-rolled specimen were broken into small ones after the tensile tests. Hypothetically, the volume of the small grain is the product of the transversal area \( A \) and grain size \( d \), and the maximum strain of the specimen is constant after tensile fracture. The deformation energy is related to SF [22], which can be expressed as

\[
E = \left( \sigma_0 + Kd^{-\frac{1}{2}} \right) \frac{\varepsilon}{V} = \frac{1}{SF} \frac{\varepsilon}{A \cdot d} = \frac{1}{SF} \times \frac{\varepsilon'}{A},
\]

where \( E \) is the local plastic strain energy, \( \varepsilon \) is the macroscopic strain and \( A \) is the transversal area of the grain.

The simplified expression of equation (5) is provided as follows:

\[
\ln E = \ln \frac{1}{SF} + \ln \frac{\varepsilon'}{A} = -\ln SF + \ln \varepsilon' - \ln A
\]

\[
(ln E)' + \frac{1}{E} = -\frac{1}{SF} - \frac{1}{A}
\]

According to the test results, the value of SF was higher than that of the transversal area \( A \). Therefore, the deformation energy can be expressed as

Figure 9. Relationship of 1/SF and 1/d^{1/2}.

Figure 10. Orientation-mapping images of the four-ply specimen: (a) Original specimen; (b) Tensile specimen.
E = \frac{SF \cdot A}{SF + A} \approx A. \quad (8)

The deformation energy increased with a decrease in grain size, and the increase in the part size corresponded to the increment of the surface energy of the small grains.

The matrix of CS near the bonding interface was ferrite [14], and carbide M_7C_3 precipitated in the interior. The binding energy of C–metal is higher than that of metal–metal, and thus, the lattice binding energy of carbides in the ferrite section is high [23]. Herein, the fragmented carbide particles after the tensile test were distributed at the bonding interface. Therefore, the deformation energy of the bonding interface was higher than that of the matrix interior and thus implied that the clad plate exhibited unstable performance at the bonding interface. In improving the bonding properties of the composite plate, a rational heat treatment process is necessary to avoid the detachment of the tensile specimen at the bonding interface.

4. Conclusions

(1) A carbide flake layer formed in the bonding interface of CS/SS. This flake easily produced stress concentrations and resulted in a fibrous region during tensile testing. The carbide layer thickness of the tensile specimen increased by approximately three times relative to that of the original specimen.

(2) The sample underwent plastic deformation via slipping during the static tension test, and this condition rapidly decreased the SF. As a result, the grain sizes ranging from 12 μm to 120 μm decreased to less than 5 μm after the tensile test.

(3) The grains broke down into small particles and extended along the slip direction during the static tension test. Carbide hardly entered the SS and only extended into the soft ferrite area of the CS surface, thereby increasing the thickness of the interface. The deformation energy increased with a decreased in grain size, and the increase in the part size was the increment of the surface energy of the small grains.

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ORCID iDs

Haibin Li https://orcid.org/0000-0002-8600-0524
Meirong Shuai https://orcid.org/0000-0001-8493-6102

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