Laser clad AlSiCuNi functionally graded coatings

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Abstract

This paper presents an exploration of laser clad AlSiCuNi-alloy FGCs on cast Al-alloy substrates. SEM microstructure observations indicate that, besides Si primary particles, hard Al3Ni2 compounds also exhibits a continuous increase in both particle sizes and volume fractions from the bottom to the top surface of the FGCs. Abrasive wear tests showed that laser clad AlSiCuNi FGCs possess much improved wear resistance and nearly the same level of friction coefficient with respect to the cast Al-alloy substrates. Tensile tests on specially designed specimens demonstrate that the interfacial bond strength between the FGCs and the substrates is higher than the ultimate tensile strength of the substrates. Digital imaging correlation (DIC) technique is employed to reveal the deformation behavior and the graded strain field of the FGCs near the interface.

1 Introduction

Laser cladding on aluminum alloys has attracted considerable interest to enhance the mechanical and chemical resistance of Al alloy components. However, a major drawback in laser cladding is the sharp interface between the coating and the substrate built up as a result of the difference in their physical and mechanical properties, which is always a potential source of weakness. This problem may be overcome if the sharp interfaces are eliminated by introducing a gradual transition from substrate to coating. Such an approach has led to the development of functionally graded coatings (FGCs) [1-3]. Our recent work was AlSi40 FGCs produced by a one-step laser cladding process on cast Al-alloy substrate [4], in which a progressive change in both microstructure and
consequent properties were achieved over the molten boundary of laser pool. The knowledge obtained from that work provides us a base to develop new FGCs with more complicated constitution according to potential applications. This paper presents an exploration of laser clad AlSiCuNi-alloy FGCs, in which both the Si primary particles and the Al$_3$Ni$_2$ intermetallic compounds exhibit a continuous increase in particle sizes and volume fractions along the thickness of the FGCs. The laser processing parameters were investigated in order to obtain a thick FGC in one-step cladding with powder dynamic feeding technology. Scanning electron microscope (SEM) observation and X-ray diffraction were employed to reveal the graded microstructure and the phases formed in the FGCs. Wear and tensile tests were performed to examine the interfacial strength and wear resistance of the FGCs. Digital imaging correlation (DIC) technique was applied to reveal the deformation behavior and the graded strain field of the FGCs near the interface.

2 Experimental Procedure

Substrates were cut from cast rods of a commercial Al-alloy with a nominal composition (wt.\%): 18.0Si, 1.1Mg, 0.3Fe, 1.1Cu, 1.1Ni and Al in balance. Standard face milling finished the surfaces of the flat substrate specimens in dimensions of 100×50×10 mm$^3$. The coating material was Al$_4$Si$_{30}$Ni$_{20}$Cu$_{8}$B$_{0.5}$ particulate, namely a mixture of Al$_{40}$Si$_{40}$ powder and a commercial Ni-alloy powder with corresponding composition.

A HAAS HL3006D-type 3kW Nd:YAG laser was used for this work. The laser beam was transported by means of a Ø 0.6mm fiber, resulting in a homogeneous intensity distribution. The focal length of the focusing lens was 140mm. It was operated at 25mm defocusing distance with a Ø 3.34mm spot on the surface of the substrate for cladding. A numerically controlled 4-axial machine executed the specimen movement. The powder feeder used was a commercial instrument (Twin-10C type) of Schulzer Metco Company. The processing parameters varied in the region of 2000~3000 W laser power, 8.3~26.7 mm/s beam speed and 10~30 g/m powder feed rate. The shielding gas was helium with a flow rate of 0.167 l/s.

Transverse sections of the clad tracks were cut for microstructural investigations. The samples were etched with 2% NaOH solution for 5~10 seconds at 40°C. A Philips XL-30 FEG scanning electron microscope (SEM) and a standard optical microscope were employed for the examination of the microstructure. The crystallographic phases of the coatings were determined with X-ray diffraction (Philips PW-1830) using Cu-K$_\alpha$ radiation. A Shimadzu HMV-2000 type micro Vickers was used for hardness measurements.
3 Results and discussions

3.1 Study of laser cladding parameters for FGCs

Figure 1a shows a series of single clad FGC tracks. The tracks have very regular contours and smooth surfaces as a result of the optimized cladding process and the use of a cooling block [4]. The specimen was mounted on the cooling block with the bottom in direct contact with flowing water maintained at a constant temperature. In this way the substrate was kept at a constant temperature as much as possible with changes of less than 5°C during laser cladding. This ensured that the geometry and dilution degree of FGC tracks depended only on the processing parameters. A cross section of the tracks is sketched in Fig. 1b, where the track performance is described with the geometric parameters of width (w), thickness (t), build-up (h) and melted depth (d). The dilution of a track by melting a thin layer of the substrate into the coating is calculated via the expression

\[ \text{Dilution} = \frac{A_1}{(A_1 + A_2)} \times 100\% \]

where \( A_1 = h(3h^2 + 4w^2)/6w \), \( A_2 = d(3d^2 + 4w^2)/6w \).

Figure 1: Geometry of laser clad AlSiCuNi FGC tracks: (a) real samples and (b) sketch of a cross section.

Figure 2 demonstrates the influence of the powder feed rate on the track performance. In the investigated range, there is a linear increase in the track thickness, width and build-up with increasing powder feed rate. It is understandable that the more powder is added the bigger the dimensions of the tracks are. However, the dilution of a track decreases dramatically with increase in powder feed rate at the beginning and the trend of further decrease slows down when the rate increases over 20g/m, as seen in Fig. 2b. Such a transition appears to be a sign that the value of the powder-feed rate is reached for a sufficient addition of the coating materials.

The effect of laser energy density on the track performance is presented in Fig. 3. The term of "laser energy density" \( (I_0) \), often referred as "laser fluence" is defined as the product of the laser power and the irradiation time on a local area divided by the beam spot area. It takes into account all the effects of laser power \( (P) \), spot size \( (D) \) and scanning speed \( (v) \) according to the equation
From Fig. 3, it is clear that the dilution and the track width increase linearly with $I_0$. On the other hand, the track thickness decreases firstly with increasing $I_0$ and keeps almost constant when $I_0$ is higher than 75 J/mm². This may be attributed to the sideways flow of the hot melt in the laser pool due to the enhanced flow away of the melt with temperature rise. Another evidence for the understanding is the continuous reduction of the track build-up with $I_0$. It is interesting to note that the track width is much larger than the beam spot size (Ø3.34-mm), which was not so obvious in AlSi40 FGCs under the same cladding conditions [4]. This is probably related to the improved wettability of the melt to the substrate due to the addition of alloying element Ni.

From the data presented above, one can see that a wide operation window of laser powder cladding could be established for high quality coatings of AlSiCuNi on the cast Al-alloys substrate. However, to obtain an ideal FGC of the same coating material with a one-step processing, the processing parameters are limited to a relatively narrow region. In general, a slightly high degree of dilution and a fast scanning speed of the laser beam are usually beneficial for the in situ formation of the gradient microstructure of the FGCs.

![Figure 2](image2.png)  
**Figure 2:** Influence of powder feed rate on the track performances: (a) thickness and width and (b) build-up and dilution. The laser power was 3000W.

![Figure 3](image3.png)  
**Figure 3:** Effect of laser energy density on the track performances: (a) thickness and width and (b) build-up and dilution. The powder feed rate was kept at a constant of 23.4g/m.
3.2 Microstructure of AlSiCuNi FGCs

Figure 4a shows a cross section of a laser-clad single AlSiCuNi FGC track. The feature of a gradual change characterizes the transition from the substrate to the FGC layer. Some rather big silicon particles of the substrate, which have not been fully melted during laser cladding, remain in a zone of a width of about 300μm over the molten boundary. The resolidified fine Si particles in the FGC layer appear at a distance of 50μm away from the boundary. This distance covers a particle-free zone, which is composed of α-Al dendrites and eutectics.

The XRD pattern in Fig.4c demonstrates at least five phases present in the AlSiCuNi FGCs, namely Al, Si, Al3Ni2, Al3Ni and nickel silicide. SEM observations combining with energy dispersive X-ray analysis (EDAX) confirm that the coarse plates possess the metastable Al3Ni2 phase rather than the stable Al3Ni phase. The latter is just formed as one of the ternary eutectics of Al, Al3Ni and Si, which are clearly revealed in Fig.5. Al3Ni2 serves to a much higher hardness than Al3Ni, 11 GPa with respect to 7.7 GPa [5], and is expected to reinforce the FGC as well. The nickel silicides are particles at sub-micrometer scale, which need further examination with transmission electron microscopy.

The gradient microstructure of the AlSiCuNi FGC is detailed in Fig.6. The FGC layer consists of Si equi-axed particles, bright plates of AlNi compounds, dark dendrites of α-Al and fine eutectics. From the bottom to the top surface of the layer, the Si particles and the Al3Ni2 plates increase in size by factors of 3.3 and 4.3, respectively. Consequently, their volume fractions increase continuously along thickness too. The detailed data are given in Table 1, in which the size of Si particles was calculated as an equivalent diameter of hexagons according to their area fraction. The morphology of the Si particles also changes accordingly from small polygons to a coarsely branched equiaxial shape.

The observed microstructural evolution inside an FGC track, i.e. more and smaller Si and Al3Ni2 particles appear in the bottom region of a track where the growth rate was slower, is apparently in contrast to the classical relation of nucleation rate-growth rate. This can be understood with a nucleation model [4] based on catalytic Si patches of different size that result from the incomplete decomposition of the original Si primary phases in the AlSi40 powder. It is suggested by this model that more Si patches with a size exceeding the critical size for heterogeneous nucleation exist in the undercooled region (the bottom region), and less in the over-heated part (the top region). The thermal gradient that corresponds to the interval between the isotherms of the liquidus of the coating alloy and of the substrate may limit the time for growth of the Si particles available at different depths and consequently dominate the final size of Si particles. Al3Ni2 plates, solidified as the second phase, nucleate readily on the Si particles and just follow a similar evolution of particle number density and size as the Si particles do with the thickness of the FGCs.

The hardness profile present in Fig.4b indicates that the FGC is 4 times harder than its substrate and, the transition between them is gradual. Some fluctuations appear on the hardness curve due to the inhomogeneity of the microstructure.
Figure 4: (a) SEM micrograph of the cross section of the AlSiCuNi FGC clad at 3000W laser power and 800 mm/min beam speed; (b) hardness distribution on the cross section and (c) XRD pattern. The unmarked peaks most likely belong to Ni$_3$Si$_2$.

Figure 5: High magnification SEM micrographs showing the main phases present in the FGCs: (a) secondary electrons image and (b) back scattered electrons image on the same field of observation.
Figure 6: SEM micrographs of the rectangular areas indicated in figure 4a showing the graded microstructure of the AlSiCuNi FGC: (a) in top region; (b) at intermediate region and (c) at the bottom.

Table 1. Size and volume fraction (V,.) of the hard phases in the AlSiCuNi FGC

<table>
<thead>
<tr>
<th>Part of the FGC</th>
<th>Si particles</th>
<th>Al$_3$Ni$_2$ plates</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Size, µm</td>
<td>V, %</td>
</tr>
<tr>
<td>Top region</td>
<td>18.61</td>
<td>29.91</td>
</tr>
<tr>
<td>Intermediate region</td>
<td>9.86</td>
<td>27.98</td>
</tr>
<tr>
<td>Bottom region</td>
<td>5.65</td>
<td>25.41</td>
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</tbody>
</table>
3.3 Interfacial bond and deformation behavior of AlSiCuNi FGCs

In order to check the bond strength between the AlSiCuNi FGCs and the cast Al-alloy substrate, a series of in-situ tensile experiments were done in the E-SEM. The laser-clad samples were specially designed, such that the FGC track just filled in the grooved substrate as a whole. After the finishing of tensile specimens, the FGC track looks like a weld located at the center of the specimen and perpendicular to the tensile loading direction, as shown in Figs. 7a and b. It ensures that the bonding interfaces between the FGC track and the substrates on both sides are under the applied load during tensile test.

All the specimens without a notch fractured in the substrate material, namely, out of the laser clad AlSi40 FGC tracks. This means that both the strength of the FGC tracks and their interfacial adhesion are higher than the ultimate tensile strength (UTS, 210 MPa) of the substrate. About 70% of the specimens fractured just about 1.5 mm away from the FGM track, as seen in Fig. 7b, and the rest experienced fracture at random positions far away from the track. However, it was found that there was no notable difference in the UTS between these two categories of the fractured specimens.

Figure 7: (a) sketch of the preparation of tensile testing samples; (b) fractured specimen showing the most probable cracking position i.e. at the boundary of HAZ and (c) a typical strain field over the FGC track mapped by DIC measurements.

In order to study the different deformation behavior of the FGC tracks with respect to the substrate, Digital Imaging Correlation (DIC) was applied [6]. A typical strain field over an FGC track is presented in Fig. 7c. The FGM track shows homogenous fields of very low strains and gradual strain transition over the interface that results to the excellent bond. In contrast, the substrate exhibits rather inhomogeneous strain fields that consist of locally heavily deformed areas,
which may develop as potential sites of failure. In particular, softening effects are clearly observed in the heat-affected zones (HAZs) of the substrate. The outer regions of the HAZs, which are adjacent to the thermally undisturbed substrates, exhibit high strain levels in the most of time during tensile tests and may initiate the final fracture in the most of cases. These observations may explain the fact that about 70% of the tensile samples fractured at the outer region of the HAZs, although its UTS reaches the same value of the substrate materials.

3.4 Wear resistance of AlSiCuNi FGCs

To characterize the wear behavior of the FGCs and of the cast Al-alloy substrate materials respectively, a pin-on-disk tribometer was employed for both sliding and abrasive wear experiments. The specimens were used as pins with a flat contact to a rubber-bond grinding disk. The abrasive wear parameters were set to 10-25 N load, 0.5-1m/s speed and a total wear distance of 3600m for each specimen. The results of wear test are promising. The AlSiCuNi FGCs exhibit slightly lower friction coefficient and 4 times higher wear resistance than the substrate materials, as seen in Fig. 8a.

![Graph showing abrasive wear rate](image1)

**Figure 8:** Results of abrasive wear experiments: (a) improved wear resistance of the FGCs with respect to the substrate materials and (b) SEM picture showing the strengthening effects of Al$_7$Ni$_2$ compound against wear.
SEM observation of worn surfaces suggests that the wear mechanisms of the FGCs and the substrate materials are quite different from each other. Under the condition of abrasive wear, cast AlSi alloys show microcracking and microcutting mechanism. In contrast, AlSiCuNi FGCs demonstrate a microcracking mechanism. As shown in Fig.8b, the Al3Ni2 intermetallic compounds greatly strengthen the FGCs and exhibit higher toughness than the primary Si particles, which leads to less microcracking of them under abrasive attack. Detailed wear tests under high load conditions are planned for better understanding of the wear mechanism of the FGCs.

4 Conclusion

The main conclusions of this work are the following:
1. Using a one-step laser powder cladding process, AlSiCuNi functionally graded coatings can be obtained that consists of silicon primary particles, Al3Ni2 plates, α-Al dendrites and ternary eutectics.
2. The Si particles and Al3Ni2 plates exhibit a continuous increase in both size and volume fraction from the bottom to the top of the FGC tracks. The morphology of the Si particles also changes correspondingly from small polygons to a coarsely branched equiaxial shape. These hard phases serves as reinforcement for the FGCs.
3. Tensile tests on specially designed specimens demonstrate that the interfacial bond strength between the FGCs and the substrates is higher than the ultimate tensile strength of the substrates. DIC measurements reveal the graded strain field of the FGCs near the interface that results to the wonderful bond. The softening effect in the heat affected zones may induce the final fracture.
4. Abrasive wear tests showed much improved wear resistance of the FGCs with respect to the substrates.

References