Friction Stir Processing of a D2 Tool Steel Layer Fabricated by Laser Cladding

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Abstract

Friction stir processing of a D2 tool steel layer fabricated by laser cladding was carried out to modify its as-cast microstructure. The microstructural evolution in the stir zone was observed and micro-Vickers hardness was measured. A maximum hardness of 857 HV was obtained by friction stir processing at 200 rpm. This value is higher than that of the clad layer when it is subjected to austenitizing heat treatment without friction stir processing. This superior property is attributed to the decrease of volume fraction of retained austenite after friction stir processing. The friction stir processing condition with low heat input led to high volume fractions of fine M_7C_3 carbide particles and martensite.

Keywords: Friction stir processing; Laser cladding; Tool steel; Hardness

1. Introduction

Much attention has been paid to laser cladding due to its superior features in the field of additive manufacturing, metallic coating, and repairing. Laser cladding equipment with a laser head, a powder feeder, and a positioning device is able to laminate metallic layers on a substrate by spraying powder and melting it with a laser beam. Fabrication of tool steel layers by laser cladding is very attractive to manufacturing high value components, such as dies and cutting tools [1-4]. However, clad layers have as-cast microstructure because of melting metallic powder with a laser beam during the process, which results in lower mechanical properties than conventional rolled metallic materials.

On the other hand, friction stir processing (FSP) has been investigated for strengthening surfaces of metallic plates. A rotating tool with a probe and a shoulder is inserted into the substrates and it produces a highly deformed zone by stirring. Fine equiaxed grains can be obtained in the stir zone because of dynamic recrystallization. FSP of metallic alloys, such as aluminum and magnesium alloys, copper alloys, stainless steel, and low carbon steel has been demonstrated [5-10]. FSP of tool steels has also been carried out to modify the surface of the plates. A stir zone with fine microstructure was produced, which resulted in a higher hardness than that of tool steels that were subjected to conventional heat treatments [11-14]. In addition, laser cladding using tool steel powder is useful for fabricating a clad layer that excludes coarse primary carbides which are generally included in rolled tool steel plates [14].

Friction stir processing seems to be an attractive approach to modify the as-cast microstructure of tool steel layers fabricated by laser cladding. However, the FSP conditions for tool steel layers after laser cladding have not been clarified sufficiently. Heat treatments are significant for obtaining fine microstructure of tool steels, and hence heat treatment conditions, such as austenitizing temperature, cooling rate, and tempering temperature, have been controlled strictly. For FSP, the heat input should be controlled to evolve the tool steel microstructures. In this study, FSP of a D2 tool steel layer produced by laser cladding was carried out to investigate the effect of FSP conditions on the microstructural evolution of the tool steel clad layer. The microstructural evolution in the stir zone was observed and micro-Vickers hardness was measured.

2. Experimental procedures

A commercial tool steel powder (American Iron and Steel Institute D2, mean particle size: 61 μm) was used for laser cladding. Its chemical composition is shown in Table 1. The combination of laser cladding and FSP is illustrated in Fig. 1. The powder was clad on a D2 tool steel plate (nominal composition: 1.40–1.60 mass% C, <0.60 mass% Mn, <0.60 mass% Si, 11.0–13.0 mass% Cr, <0.30 mass% Ni, 0.70–1.20 mass% Mo, <1.10 mass% V, balance Fe) with laser cladding equipment that consisted of a coaxial cladding head (Laserline, Type Coax 8), a diode laser system (Laserline, LDF-1000-750), and a powder feeder. The position of the cladding head was controlled with a

robotic arm (Yasukawa, Motoman). The tool steel layer (65 mm × 15 mm × 2.5 mm) was formed on the substrate by multi-pass laser cladding at a laser power of 600 W. The gap between the cladding paths was 0.5 mm in the transverse and vertical directions. The scanning rate of the laser beam, the beam diameter at the plate surface, and the powder feeding rate were 1000 mm/min, 1 mm, and 10 g/min, respectively. Argon gas was used as the carrier gas for the powder (3 L/min) and as a shielding gas (20 L/min). The melt-in depth of the upper part of the substrate was about 40 micrometers. The clad layer was bonded to the substrate without any defects. The top surface of the clad layer was polished using a grinding machine to form a flat surface before FSP.

The obtained clad layer was subsequently modified by FSP. An FSP tool made of hard metal (WC-Co alloy) with a columnar shape (12 mm diameter) and an unthreaded probe (7 mm diameter and 0.7 mm height) was rotated from 200 to 600 rpm and travelled from 50 mm/min to 400 mm/min. The shape of the shoulder was concave. A tool tilt angle of 3° was used. FSP forging load was changed from 19.6 kN to 39.2 kN. The processed area was shielded with argon gas during FSP (20 L/min). The temperature 1 mm below the process centerline was measured using an embedded K-type thermocouple (0.5 mm diameter). To compare results with conventional heat treatment, the clad layer was subjected to austenitizing heat treatment without FSP. The heat treatment comprised austenitizing the samples at 950°C to 1050°C for 600 s in air in an electric furnace and cooling them in air.

The microstructure of the clad layer was observed with an optical microscope (Nikon, EPIPHOT-TME) and a scanning electron microscope (SEM, JEOL, JSM-6610LA). The samples that had been subjected to FSP were cross-sectioned perpendicular to the travelling direction, polished mechanically, and etched with a 5% nitric acid and methanol solution. The area fraction of carbides on each of the SEM images was determined by image analysis. Crystal structures of the matrix and the carbide particles were identified by X-ray diffractometry (XRD, Rigaku, RINT2500V). The diffractometer was operated at 40 kV and 40 mA using Cu-K α radiation ($\lambda = 0.154$ nm). The volume fraction of retained austenite was determined by comparing the relative intensities of the ferrite (α_{200} , α_{211}) and austenite (γ_{200} , γ_{220}) peaks. The microhardness along the centerline of the cross section in the stir zone was measured using a micro-Vickers hardness tester (Akashi, HM-124) with a load of 300 gf.

3. Results and discussion

3.1. Laser cladding and friction stir processing

As-received D2 tool steel powder contained a single austenite phase, as shown in Fig. 2. All alloy element atoms, i.e., carbon, chromium, molybdenum, and vanadium, were therefore trapped in the austenite. The clad layer was fabricated successfully by laser cladding. Fig. 3 shows the microstructure and XRD pattern of the clad layer. Dendrites and carbides in the interdendritic region

were observed in the cross section of the clad layer. This indicates that the tool steel powder melted and then solidified during the laser cladding process. The XRD pattern revealed that the structure of the clad layer consisted of martensite, M_7C_3 , and retained austenite. The volume fraction of retained austenite and M_7C_3 after laser cladding was 63% and 5.2%, respectively. It is thought that martensite was formed during cooling due to the decrease of the carbon content in austenite by laser cladding. The hardness of the clad layer was 478 HV.

Table 2 shows the surface hardness of the clad layer after FSP under various conditions. FSP under the condition (a) produced hard layer without any defects. The effect of the travelling speed on the surface hardness was relatively small (conditions (b) and (c)). When the travelling speed was 400 mm/min (condition (d)), the rotating tool fractured during FSP due to low heat input. Higher forging load (condition (e)) also caused fracture of the rotating tool. When the forging load was 19.6 kN (condition (f)), the rotating tool could not be plunged into the clad layer. Higher rotation speed of 400 rpm (condition (g)) caused groove-like defects in the clad layer. When the rotation speed was increased (conditions (h) and (i)), the surface hardness was considerably decreased, compared with the condition (a). As the rotation speed significantly influenced the surface hardness, the effect of the rotation speed is described below.

The cross section of the clad layer modified by FSP at 200 rpm is shown in Fig. 4. No defects were observed in the stir zone. The depth of the stir zone corresponds to the probe length of the

rotating tool.

The cooling rate during FSP was insensitive to the rotation speed, as shown in Fig. 5. The average cooling rate from 950°C to 250 °C was 25 °C/s, 26 °C/s, and 30 °C/s when the rotation speed was 200, 400, and 600 rpm. All cooling rates were sufficient to suppress diffusion-type phase transformations, compared with the upper critical cooling rate (0.9 °C/s) for D2 tool steel [15]. The peak temperature during FSP increased with increasing rotation speed. At a rotation speed of 200 rpm, the peak temperature was 959°C, which was the lowest among these samples.

3.2. Microstructural evolution of clad layer by FSP

Back-scattered electron images and hardness of cross sections of clad layers subjected to austenitizing heat treatment without FSP are shown in Fig. 6. The austenitizing heat treatment after laser cladding hardly modified the dendritic microstructure. Carbide particles precipitated mainly on the boundaries of austenite dendrites. The area fraction of carbide particles in the clad layer austenitized at 950°C, 1000°C, and 1050°C was 12.6%, 10.6%, and 7.5%, respectively. The respective volume fraction of austenite was 9.5%, 10%, and 12%. The decrease in austenitizing temperature led to larger amounts of carbide particles, smaller amounts of retained austenite, and a higher hardness.

Fig. 7 shows the depth profiles of the hardness distribution of the FSPed clad layer. The hardness

in the stir zone increased with a decrease in tool rotation speed. The highest hardness of 857 HV was obtained by FSP at 200 rpm, which was superior to that of the clad layer subjected to austenitizing heat treatment. Based on the temperature measurement in Fig. 5, cooling rates were independent of tool rotation speed. Therefore it is thought that the variation in hardness depends on the peak temperature during FSP. The hardness in the stir zone FSPed at 400 and 600 rpm was relatively low and dependent on the distance from the top surface. This means that the lower heat input leads to a higher hardness. These results correspond with the relationship between austenitizing temperature and D2 tool steel hardness in conventional heat treatment.

The XRD analysis at 0.5 mm below the top surface of the stir zone is shown in Fig. 8. The amount of martensite increased and weak peaks of retained austenite were present after FSP at 200 rpm. The volume fraction of retained austenite in the clad layer FSPed at 200, 400, and 600 rpm was 5.7%, 17%, and 30%, respectively. The structure of the stir zone FSPed at 400 and 600 rpm contains considerable amounts of retained austenite. The decrease of volume fraction of retained austenite could contribute to a hardness that is higher than that of the clad layer subjected to austenitizing heat treatment without FSP.

Secondary electron images of cross sections of the stir zone FSPed at 200 rpm are shown in Fig. 9. They are changed significantly by FSP, compared with the microstructure in the clad layer without FSP (Fig. 6). Fine equiaxial prior-austenite grains were obtained after FSP. Fine carbide particles

indicated by arrows were dispersed uniformly in the stir zone. They precipitated both on the grain boundaries and within the grains. Elemental analysis by SEM-energy dispersive X-ray spectroscopy revealed that the carbide particle was comprised of 32.4 at%C, 22.7 at%Cr, 43.8 at%Fe, 0.57 at%V, and 0.53 at%Mo, which corresponded to M_7C_3 . The size of carbide particles was reduced by increasing the distance from the top surface owing to lower heat input. At 0.8 mm below the top surface of the stir zone FSPed at 200 rpm, the carbide particle sizes and prior austenite grains were <400 nm and <1 μ m, respectively.

Cross sections of the stir zone FSPed at 400 rpm were shown in Fig. 10. The microstructures were coarser than those in the stir zone FSPed at 200 rpm. The amount of carbide particles increased with increasing distance from the top surface.

The area fraction of carbide, mean size of carbide, and mean size of austenite grains after FSP is shown in Fig. 11. The amount of carbide particles in the stir zone FSPed at 200 rpm was higher than that of the stir zone FSPed at 400 and 600 rpm. The amount of carbide particles increased with increasing distance from the top surface. Consequently, the FSP condition with low heat input leads to a high volume fraction of carbide particles. The amount of carbide particles in the stir zone FSPed at 200 rpm was higher than that of the clad layer austenitized at 950°C, 1000°C, and 1050°C without FSP (12.6%, 10.6%, and 7.5%). This implies that preferential nucleation sites for precipitation of carbide particles should be increased by the stirring effect. The size of carbide particles and austenite

grains in the stir zone FSPed at 200 rpm was lower than that of the stir zone FSPed at 400 and 600 rpm. The size of carbide particles and austenite grains decreased slightly with increasing distance from the top surface.

When the clad layer was subjected to heat treatment at 950°C, 1000°C, and 1050°C, the variation of the amount of retained austenite was relatively small (9.5%, 10%, and 12%). In contrast, when the rotation speed of FSP was changed, the amount of retained austenite was considerably varied (5.7%, 17%, and 30%). Fujii et al. reported that severe plastic deformation during friction stir welding stabilized austenite phase in steel joints [16]. Plastic deformation creates a high density of dislocations, which increases the magnitude of the internal stress of the austenite phase. It is thought that austenite was stabilized by a high density of dislocations when the rotation speed was 400 and 600 rpm in this study. On the other hand, when the rotation speed was 200 rpm, the volume fraction of carbide particles was significantly high and almost constant irrespective of the depth as shown in Fig. 11. It is thought that large amounts of carbide particles precipitated on the dislocations acting as preferential nucleation sites during FSP. The precipitation reduced the dislocation density and carbon content in austenite and prevented austenite from being stabilized, which resulted in high hardness in the stir zone. Meanwhile, the hardness depth profiles showed the peak at 1 mm depth when the rotation speed was 400 and 600 rpm. It is thought that lower heat input at the bottom of the stir zone increased the amount of carbide precipitates, and reduced the dislocation density and the amount of

retained austenite.

Conclusions

FSP of a clad layer fabricated by laser cladding using D2 tool steel powder was carried out. The microstructural evolution in the stir zone was observed and micro-Vickers hardness was measured.

The results are summarized as follows:

(a) Formation of carbide particles in the interdendritic region and martensite transformation in the matrix occur during the laser cladding process. Subsequent FSP produces a homogeneous distribution of fine M_7C_3 carbide particles and increases the amount of martensite.

(b) The highest hardness of 857 HV is obtained by FSP at 200 rpm, which is higher than that of the clad layer subjected to austenitizing heat treatment. This superior property is attributed to the decrease of volume fraction of retained austenite after FSP.

(c) The FSP condition with low heat input leads to high volume fractions of fine M_7C_3 carbide particles and martensite.

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Captions

Fig. 1 Schematic illustration of processes: (a) laser cladding and (b) subsequent friction stir processing.

Fig. 2 X-ray diffraction pattern of as-received D2 tool steel powder.

Fig. 3 Clad layer of D2 tool steel after laser cladding process: (a) cross section and (b) X-ray diffraction pattern.

Fig. 4 Cross section of clad layer after friction stir processing at 200 rpm.

Fig. 5 Temperature variation measured 1 mm below top surface of clad layer during FSP: (a) at 200 rpm, (b) at 400 rpm, and (c) at 600 rpm.

Fig. 6 Cross section of clad layer after austenitizing: (a) at 950°C, (b) at 1000°C, and (c) at 1050°C.

Fig. 7 Depth profile of hardness distribution of FSPed clad layer.

Fig. 8 XRD patterns of FSPed clad layers 0.5 mm below top surface: (a) at 200 rpm, (b) at 400 rpm, and (c) at 600 rpm.

Fig. 9 Cross section of clad layer after FSP at 200 rpm: (a) 0.1 mm depth, (b) 0.5 mm depth, and (c) 0.8 mm depth.

Fig. 10 Cross section of clad layer after FSP at 400 rpm: (a) 0.1 mm depth, (b) 0.5 mm depth, and (c) 0.8 mm depth.

Fig. 11 Area fraction of carbide, mean size of carbide, and mean size of γ grains after FSP at 200, 400, and 600 rpm.

Table 1 Chemical composition of as-received D2 tool steel powder (mass%).

Table 2 Surface hardness of clad layer after FSP under various conditions.

Table 1 Chemical composition of as-received D2 tool steel powder (mass%).

C	Si	Mn	Cr	Mo	Al	V	Fe
1.50	0.28	0.48	11.0	0.83	0.005	0.36	Bal.

Table 2 Surface hardness of the clad layer after FSP under various conditions.

Condition	Rotation speed (rpm)	Travelling speed (mm/min)	Load (kN)	Hardness (HV)
(a)	200	50	32.3	796
(b)	200	100	32.3	801
(c)	200	200	32.3	754
(d)	200	400	32.3	No Data
(e)	200	50	39.2	No Data
(f)	200	50	19.6	No Data
(g)	400	50	19.6	No Data
(h)	400	50	32.3	602
(i)	600	50	32.3	514

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Figure

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