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Interfacial bonding features of Ni coating on Al substrate with different surface pretreatments in cold spray

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ABSTRACT

The effect of the substrate surface pretreatment on the Ni–Al interfacial bonding features in cold spray was investigated. Oxide-free interfacial area is found to be the essential factor that affects the interfacial atomic diffusion and metallurgical bonding. Voids are formed at the interface between the particle and grit-blasted substrates, significantly reducing the oxide-free interfacial area. However, those voids cannot be found at the interface between the particle and polished and ground substrates. Therefore, the polished and ground substrates can provide higher coating mass and coating-substrate bonding strength than the grit-blasted ones.

1. Introduction

Cold spray is a relatively new coating technique in which the powders are accelerated to a high velocity in a supersonic jet and impact on the substrate or already deposited coating at an entirely solid state. The investigation on the bonding mechanism of the cold sprayed coating is always a hot topic. Currently, the most acceptable view can be regarded as the adiabatic shear instability (ASI) at the localized interfacial region where the thermal softening is dominant over the work hardening to form the outward metal jet [1,2]. Such viscous-like metal jet helps to clean up the cracked native oxide film which originally exists on the particle and substrate surfaces, allowing the oxide-free contact and thus the metallic bonding to occur [3]. Metallurgical bonding and mechanical interlocking are commonly perceived to be two mechanisms of the metallic bonding in cold spray. Metallurgical bonding is known as a result of the atomic diffusion at the oxide-free interface. Basically, the formation of intermetallic phase at the interfacial region [4–11], dimple-like and groove-like features appearing at the fracture surface [12–14] or very thin amorphous layer [15,16] can be recognized as a marker for oxide-free interface and thus true the metallurgical bonding. As for the mechanical interlock, it results from the extrusion and physical mixture of the soft metal materials and always happens at the coating-substrate interface. The strength of mechanical-based bonding is normally much smaller compared with that of metallurgical-based bonding.

Particle–particle bonding is only determined by the original properties of the feedstock and the working parameters, while for the coating-substrate bonding, the substrate surface conditions make the bonding process to be more complicate. Therefore, many efforts have been devoted to study the effect of substrate surface conditions on the coating-substrate bonding during the past years. Some experimental works reported that grit-blasted substrate results in slightly higher DE than polished one [17,18], and coarse-grit-blasted substrate gives a coating mass of 10% higher than fine-grit-blasted one [19]. However, a number of works pointed out that polished or ground substrate significantly increases the coating-substrate bonding strength in comparison with grit-blasted substrate [20–22]. Although the explanations on those measuring data were provided in their papers, the convincing experimental evidences to support their hypothesis were still lacking. In addition, Kumar et al. reported a contradictory result that the bonding strength (Cu on Cu, Cu on Al) for the polished case is lower than that for the grit-blasted case, making this issue to be more confusing [18]. Thus, it is of particular importance to clarify what really happens in the coating–substrate interface under different substrate surface conditions. In the current work, a comprehensive investigation was performed to clarify the effect of the pretreatment method on the Ni–Al interfacial bonding features. The substrate surfaces were respectively polished, ground, blasted by small and large grits before spraying. Single particle deposition was studied experimentally and numerically to observe the interfacial features between the particle and substrate. Coating was also deposited and

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annealed to observe the formation of intermetallic phase and then to evaluate the metallurgical bonding quality at the coating-substrate interface.

2. Experimental and numerical descriptions

Coating was deposited by using a home-made cold spray system (LERMPS, UTBM, France) with the commercially available MOC nozzle whose dimensions can be found elsewhere [23]. Compressed air was used as the main driving gas with the working pressure of 2.5 MPa. A low gun traverse speed of 100 mm/s was used for the full coating deposition and each sample was coated for three times to guarantee a sufficiently thick coating. The single particle splat was deposited at the gun traverse speed of 500 mm/s. The standoff distance from the nozzle exit to the substrate surface was 30 mm. The pure Ni powder (ECKA Granules Metal Powders Ltd, Germany) with spherical morphology and the size range between 10 and 45 μm was selected as the feedstock. The pure Al bulk material with cylinder shape and same size was chosen as the substrate. The substrate surface was pretreated by four different ways and the roughness of the pretreated substrate surfaces was measured by a surface roughness tester (SJ-210, Mitutoyo, USA). The detailed pretreatment procedure and measured surface roughness are listed in Table 1. The cross-sections of the substrate surface region were observed by an optical microscope (OM) (Nikon, Japan). Fig. 1 shows the OM image of the substrate cross-section with different surface pretreatments. For evaluating the diffusion at the Ni–Al coating-substrate interface, all samples were annealed at the same condition to form the intermetallic phase. The samples were heated at 20 °C/min to the annealing temperature of 400 °C, held at this temperature for 15 min and then furnace cooled to the room temperature. The furnace cooled down at a rate of approximately 10 °C/min from the annealing temperature of 400 °C to 350 °C. The coating and splat microstructures were examined by a scanning electron microscope (SEM) (JSM5800LV, JEOL, Japan).

FEA was also performed as an assistant method to explain the experimental results. Simulation was performed by ABAQUS with the Lagrangian algorithm and Dynamic-Explicit solver [24]. The three-dimensional impacting process was assumed to be an adiabatic process and simplified as a 1/4 symmetrical model to save the computational time. The particle diameter of 25 μm was employed to represent an average level of the feedstock diameters and the other diameters were neglected because it is impossible to account for all the diameters in the simulation. But it should be noted that the particle size indeed has some effects on the particle deposition because the particle impact velocity and critical velocity are both size-dependent parameters [23]. The substrate surface was treated in different ways depending on the pretreatment method. The detailed description on the surface treatments can be found in Table 1. Fixed boundary condition was imposed to the substrate bottom and side walls. Based on the real working conditions and nozzle geometry, the particle impact velocity and temperature were calculated as 518 m/s and 31 °C by the computational fluid dynamic (CFD) technique. The substrate temperature was given as 25 °C. The materials were described by the Johnson–Cook plasticity model which accounts for strain, strain rate hardening, as well as thermal softening [25]. The property parameters for Ni and Al used in the model can be found elsewhere [26].

3. Results and discussion

Fig. 2 shows the cross-sectional SEM image of a single Ni particle deposition on different Al substrates at the working temperature of 600 °C. It is clearly observed that the particle tightly bonds with the substrate without any obvious voids at the interface for C1 and C2. As for C3 and C4, however, some voids marked by the white arrows appear at the contact interface, significantly reducing the interfacial area. The reason for this fact can be found from Fig. 1 which shows the surface conditions of different substrates. It is seen that there exist a large number of scattered micropits on the substrate surface resulting from the grit-blast for C3 and C4. When particles impact on these pits, they cannot fully contact with the exposed substrate surface but be trapped or hindered by these pits to form a bridge-like structure as indicted in Fig. 2c and d. As a consequence, the real oxide-free contact area is insufficient to guarantee the strong metallurgical bonding; most particles will rebound after impacting on the substrate. Actually, we found that the amount of the successful deposition particles for C1 and C2 is indeed much larger than that for C3 and C4 by observing the cross-section of each sample. Furthermore, in Fig. 2d concerning about C4, some defects can be recognized near the lower interfacial region. These defects arise

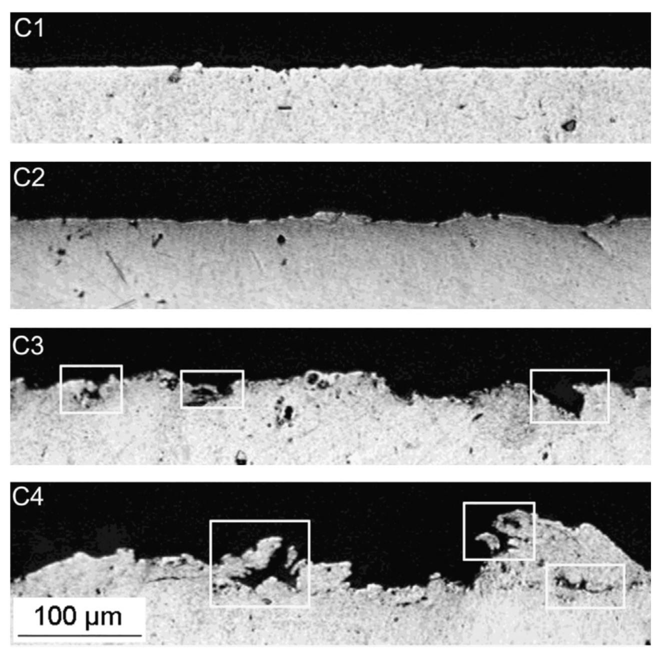

Fig. 1. Cross-sectional OM image of the substrate surface region after different surface pretreatments. C1 refers to polished substrate, C2 refers to ground substrate, C3 refers to grit-blasted substrate by small size Al_2O_3 sand and C4 refers to grit-blasted substrate by large size Al_2O_3 sand.

Table 1
Detailed description of different pretreatment methods.

	Pretreatment	Roughness	Experimental description	Numerical description
C1	Polish	0.3695 ± 0.149 μm	Grounded by SiC sand papers and then polished by 0.05 μm Al_2O_3 solution	Flat surface
C2	Ground	2.1201 ± 0.497 μm	Grounded by 200 μm SiC sand paper	Surface with circle annular crater (2 μm)
C3	Grit-blast	2.5269 ± 0.492 μm	Grit-blasted by small size Al_2O_3 sand	Surface with cylindrical crater (2.5 μm)
C4	Grit-blast	6.3545 ± 1.230 μm	Grit-blasted by large size Al_2O_3 sand	Surface with cylindrical crater (6.5 μm)