Investigation of the microstructure and mechanical performance of bimetal components fabricated using CMT-based wire arc additive manufacturing

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Abstract

Traditionally, wear-resistant components are manufactured by cladding hard facing material on the base metal. This production process is typically complicated, expensive, and time-consuming. This study proposes a method of fabricating components with high wear resistance requirements utilizing cold metal transfer based wire and arc additive manufacturing with hard facing welding wire as the consumable material. Thin-walled and block components were manufactured by depositing a combination of a low alloy steel, ER80S-G, and a hard facing material, MF6–55GP. Microstructure characterization and mechanical properties (hardness, tensile and Block-on-Ring wear test) were performed. The results revealed that the ER80S-G/MF6–55GP bimetal components were able to be fused with no detectable defects near the border. As the deposited height was increased, the residual stress also increased; this internal residual stress combined with the external tensile load lead to a very low tensile strength of 447.79 ± 24.32 MPa of the ER80S-G/MF6–55GP/ER80S-G sandwich structure. The microstructures, constituent phases, and hardness distributions differ greatly among the layers due to their different thermal histories. The wear weight loss varies as the load condition changes for both the MF6–55G and Cr12MoV steels. Compared to Cr12MoV, MF6–55GP weld metal exhibits better wear resistance at higher loads in dry sliding wear tests.

Nomenclature

CMT cold-metal transition
WAAM Wire arc additive manufacturing
AM Additive manufacturing
FGM Functional Graded Materials
LAM Laser Additive Manufacturing
SAW Submerged Arc Welding
EPMA Electron Probe Microanalyzer

1. Introduction

Additive Manufacturing (AM) has become a commonly used process for producing prototypes, components with complex geometries, and personal products. Compared to conventional subtractive manufacturing methods, AM is based on melting of powders or wires using a focused heat source, to form condensed parts using layer accumulation [1].

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While several heat sources are available in AM, lasers are the most popular choice. Laser additive manufacturing (LAM), also characterized as ‘powder metallurgy’, has been widely developed and commercialized for certain industrial applications [2]. It offers numerous advantages, including providing the ability to manufacture near-net shape components, produce functionally graded materials (FGMs), etc. However, there are also some limitations, such as the high cost of the equipment, high maintenance costs, powder waste, high porosity, oxidation issues, and low production efficiency [3].

Wire and Arc Additive Manufacturing (WAAM) attracts more than more attention due to the increasing demand for obtaining lower production cost and higher deposition efficiency [4]. Since 1980s, submerged arc welding (SAW) was firstly used to fabricate large metal parts through the accumulation of welds, substantial work has been done to develop this technology. Cold metal transfer (CMT) is one of the most widely utilized methods for WAAM. The CMT process with the advantages of very low thermal input, spatter free droplet transfer and high deposition rate, has been widely adopted for welding thin plates, aluminum and titanium alloys [5]. During the droplet transfer process, once the droplet touches the welding pool, a short circuit is detected and the current is subsequently reduced to a significantly lower level, at the same time, the welding wire is simultaneously retracted from the droplet. Cong et al. [6] studied the WAAM process using four types of CMT techniques, namely, conventional CMT, CMT pulse, CMT advanced, and CMT pulse advanced. According to their results, CMT pulse advanced process is the most suitable one for manufacturing Al-6.3%Cu alloy with almost no porosity in the as-built object. Ola et al. [7] confirmed the ability to use CMT to produce defect-free, low-dilution cladding on Inconel® 718 superalloy. Ge et al. [8] conducted preliminary studies on the possibility of producing near-net-shape H13 die steel using a WAAM process based on CMT. The internal 3D pore distribution, microstructures, and the mechanical properties of the deposited H13 blocks were investigated.

As a low-cost and highly efficient method for manufacturing large components, WAAM enables the deposition of numerous materials, such as aluminum alloys and titanium alloys, nickel alloys, stainless steels and duplex stainless steel [9]. Manufacturing of hard-facing materials is comparatively rare. Hard facing materials are mainly used as claddings to improve the hardness and wear resistance of mechanical components [10]. Due to the large material property difference between the core material and the cladding material, an intermediate layer between them is necessary to relieve the interface stress [11]. In general, the cladding process is typically complicated, expensive, and time-consuming. Therefore, a simple and efficient method of fabricating integrated wear-resistant components need to be developed.

This paper proposes a novel method of fabricating high hardness and wear-resistant parts using CMT-based WAAM. MF6-55GP, one of the most widely used iron-based hard facing materials, was chosen as the cladding layer. A large amount of Cr is alloyed in the crystal lattice, which improves the hardenability, hardness, wear and corrosion resistance, and even yield strength [12, 13]. Due to the high hardness, MF6-55GP metal can only be machined by grinding, which increases the processing difficulty. Moreover, as the deposition layer number increases, the stress concentration increases. According to the study conducted by Nie [14], the maximum residual stress is located between the as-built object and the substrate. In this study, several layers of relatively soft low-carbon steel ER80S-G were deposited between the substrate and hard facing material to decrease the local stress. Investigation of the microstructure, as well as the mechanical tests of the hardness, tensile strength, and wear of the deposited bimetal structure were carried out.

2. Experimental details

2.1. Experimental setup
The schematic diagram of the arc-based AM process used in this study is presented in figure 1. This configuration consists of a TransPuls Synergic 4000 CMT power source from Fronius and a six-axis ABB robot. Low-carbon steel (Q235) plates, with dimensions of 200 mm × 200 mm × 5 mm, were used as single use substrates and were screwed to a water-cooled aluminum alloy block. This setup was intended to achieve a faster cooling rate than using the Q235 plate as the substrate only. The welding parameters are listed in table 1.

Two commercial welding wires consisting of ER80S-G and MF6-55GP with diameters of 1.2 mm were used as the filler metals. Their nominal chemical compositions are listed in table 2. The MF6-55GP wire has a relatively high carbon content and is commonly used for repairing dies and/or molds.

2.2. Microstructure, constituent phase, and microhardness
Analyses of the microstructure, constituent phase, and microhardness were carried out on a single sample, which was prepared by depositing four single-pass layers of ER80S-G followed by four single-pass layers of MF6-55GP on a Q235 plate. Then, the sample was vertically cross-sectioned relative to the substrate, as shown in figure 2. The sample was subsequently ground, polished, and etched using C₂H₅OH-4%HNO₃ acid. An optical microscope (Olympus XJP-300) and a scanning electron microscope (SEM, FEI Nova NanoSEM 450) equipped...
with an energy dispersive spectroscopy (EDS, Oxford X-Max®) were employed to observe the microstructures. The phase composition was determined by x-ray diffraction (XRD) analysis using Bragg Brentano camera geometry and Cu Kα incident radiation.

A Vickers hardness test was employed to measure the microhardness with a load of 1.96 N for 10 s. The test was performed starting from the bottom of the sample and proceeding upwards (from ER80S-G and MF6-55GP), the horizontal distance between two test points was 1 mm, and three points were tested for each deposition layer.

### Table 1. Deposition process parameters.

| Parameters                  | Value      |
|-----------------------------|------------|
| Current DC (A)              | 82         |
| Voltage (V)                 | 12.6       |
| Travel speed (mm s⁻¹)       | 5          |
| Wire feed speed (m min⁻¹)   | 5          |
| Interpass temperature (°C)  | 130        |
| Shield gas                  | 80% Ar and 20% CO₂ |
| Shield gas flow rate (l min⁻¹) | 15       |

### Table 2. Nominal chemical composition of experimental materials (wt.%).

|       | C   | Si  | Mn  | Ti  | Cu  | Cr  | V   | Mo  | Fe   |
|-------|-----|-----|-----|-----|-----|-----|-----|-----|------|
| Q235  | 0.12–0.2 | 0.025 | 0.3–0.7 | —   | —   | 0.3 | —   | —   | balanced |
| ER80S-G | 0.08 | 0.65 | 1.25 | 0.15 | 0.20 | 1.25 | 0.26 | 0.58 | balanced |
| MF6–55GP | 0.67 | 0.70 | 0.70 | —   | —   | 5.20 | —   | 1   | balanced |
| Cr12MoV | 1.67 | 0.35 | 0.35 | —   | —   | 12  | 0.40 | 0.80 | balanced |

2.3. Tensile test

The samples for the tensile test were cut from the single-pass multi-layer thin-walled structure consisting of a combination of ER80S-G and MF6-55GP, according to the ASTM E8/E8M-09 standard. Due to the extremely high hardness of the MF6-55GP metal, it is difficult to grasp in the machine fixtures. Therefore, 20 and 30 layers of ER80S-G were deposited on the bottom and top sections of the thin-walled structure, respectively; 10 layers of MF6-55GP were sandwiched between the two ER80S-G weld metals, as shown in figure 3(a)). In this configuration, two ends of the sample are composed of the relatively mild ER80S-G, allowing them to be grasped...
by the fixtures. A mechanical testing machine (CMT5305, MTS Industrial Systems (China) Co., Ltd.) was used to perform the tensile test in displacement control with a cross head speed of 0.05 mm s\(^{-1}\).

2.4. Wear test
A Block-on-Ring rotational tribometer was used to evaluate the wear properties of the MF6-55GP weld metal under dry conditions (without lubrication), as shown in figure 4. Samples with dimension of \(\phi 10\times10\) mm were cut from the as-built multi-pass multi-layer component and mounted onto a fixed shaft; GCr15 steel (\(\phi 40\) mm in diameter and approximately 60–62 HRC) was used as a standard counter specimen and assembled on a rotating shaft. Mutual friction was generated between MF6-55GP and GCr15 under a normal force. The parameters of the wear test are presented in table 3. A commercial die steel, Cr12MoV, was also tested using the same parameters in order to make a comparison with the MF6-55GP weld metal. Its chemical composition is listed in table 1. The wear test was performed using an MMS-2A machine fabricated by Jinan Yihua Tribology Testing Technology Co., Ltd., China.
3. Results and discussion

3.1. Microstructure, constituent phase, and microhardness

Figures 5–7 show the microstructures of the as-built component shown in figure 2. Figures 5(a)–5(d) show the microstructures of the bottom four layers of ER80S-G, figures 7(a)–7(d) show the microstructures of the upper four layers of MF6-55GP, and figure 6 displays the interface between the ER80S-G and MF6-55GP. The microstructures of the ER80S-G layers are mainly consisting of ferrite and small amounts of barite and pearlite. There are no significant differences in the microstructures of these four layers; however, several morphological discrepancies exist. As shown in figure 5(a), the microstructure of the first layer mainly consists of acicular ferrite, much finer microstructure than the other three layers. This morphology difference was reflected on microhardness. The micro-hardness distribution along the vertical direction of the substrate is shown in figure 8. As shown in the figure, the hardness of the first layer was about 250 Hv, and the hardness of 2nd, 3rd and 4th layer were almost constant and slightly lower than the 1st layer. As the first layer, which is deposited directly on the substrate, heat can be rapidly conducted to the substrate (which acts as a large heat sink for the as-built object). For the subsequent layers, the higher depositing layer, the larger distance to the substrate, then the smaller temperature gradient and lower cooling rate, finally the coarser microstructure obtained. Compared to the fine acicular ferrite of the first layer, the microstructures of the 2nd, 3rd, and 4th layers were composed of mainly quasi-polygonal ferrite; these coarser microstructures led to a lower hardness of approximately 220 Hv.

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Figure 6 shows optical images of the morphology near the interface between the EG80S-G and MF6-55GP weld metal. Although these two materials are significantly different in their chemical compositions, they were able to be fused with no detectable defects near the border. No pores, cracks, or delineations were observed at this magnified scale, which indicates good metallurgical bonding. The interface between ER80S-G and MF6-55GP is wavy in shape similar to other additively manufactured bimetals. Due to the low carbon content of the 80S-G weld metal, no carbide or intermetallic compound was precipitated in this observed range [15]. According to Shakerin et al., [16] when two dissimilar materials are bonding together, an abrupt change of grain morphology, solidification structure, and chemical composition across their interface can reduce fatigue life and even corrosion resistance of as-built bimetal structure. Further study needs to be developed on the bonding mechanism between the two dissimilar materials with some innovative methods.

Table 3. Wear test parameters.

| Wear parameters | Condition       |
|-----------------|-----------------|
| Sliding conditions       | Dry sliding     |
| Sliding speed (r min⁻¹) | 200             |
| Load (N)               | 100/200/300     |
| Sliding time (min)      | 60              |
| Temperature (°C)        | Room temperature|

Figure 4. Schematic illustration of Block-on-Ring rotational wear test.
In the transition zone, between the 4th and 5th layers, the hardness increases rapidly from 220 Hv to 500 Hv. This can be explained by the differences in carbon content and microstructures between MF6-55GP and ER80S-G.

The four layers of MF6-55GP weld metal exhibited significantly more varied microstructures and morphologies than the layers of ER80S-G, which is a result of the great difference in thermal history between layers, as shown in figure 7. The 5th layer consisted of mainly tempered martensite (dark areas) and some residual austenite (bright areas). The 6th layer consisted of a combination of tempered martensite, residual austenite, and carbides. The 7th and 8th layers are composed of mainly quenched martensite, residual austenite, and carbides. It should be noticed that, figure 7(c) only presents the microstructure of the top part of the 7th layer. And this part was remelted and cooled along with the 8th layer, meaning they have very similar microstructures. As the final layer of the structure, the 8th layer (as well as the top part of 7th layer) underwent a quenching process with no post-heat treatment and exhibited a large proportion of quenched microstructure. As seen in figures 7(a)–7(d), as the layer number increases, the proportion of the tempered microstructure (dark areas) decreases, and the proportions of residual austenite and carbides (bright areas) increase. These differences in the microstructure can be used to explain the hardness distribution in MF6-55GP weld metal. For a given chemical composition, the microstructure when tempered tends to be softer than when quenched, so the
hardness continues to increase from the 5th layer to the 7th layer, and subsequently reaches a maximum value of 800 Hv in the 8th layer, which had the highest proportion of quenched microstructure. Figure 9 shows the scan of the Electron Probe Microanalyzer (EPMA) line (highlighted in figure 2), perpendicular to the interface between the ER80S-G and MF6-55GP. Changes in the amounts of Cr and Fe were detected in the observed range of 0–1600 μm. Due to the low accuracy of the EDS analysis for light elements, the distribution of carbon is not included in this scan. One thing worth to be noticed is that abrupt change in elements distribution and element diffusion along the interface can severely deteriorate the performance of bimetal components in service conditions, so deposition strategies need to be put forward to suppress the deterioration of mechanical properties near the interface.

Figure 7. Optical micrographs of the MF6-55GP weld metal.

Figure 8. Vickers hardness distribution of the thin-walled structure, oriented vertically.
The XRD patterns and phase identification of the thin-walled structure are presented in figure 10. The graphs for both the ER80S-G and MF6-55GP weld metals consist of three well-defined peaks for the $\alpha$-Fe phase at 44.8°, 65.1°, and 82.3°, corresponding to the crystallographic planes (110), (200), and (211), respectively. The $\alpha$-Fe phases in the ER80S-G and MF6-55GP were determined to be the ferrite and martensite phases, respectively. According to the analysis via optical microscope, there are some residual austenitic and carbides present in the MF6-55GP weld metal, but they were not detected by XRD due to their low amount.

3.2. Tensile test
The engineering stress–strain curves and the macroscopic picture of the fractured samples are presented in figure 11. The ultimate tensile stresses (UTS) of the three samples were 427.29 MPa, 441.41 MPa, and 474.66 MPa, respectively. The average apparent tensile strength was 447.79 ± 24.32 MPa. It can be seen that the tensile specimens made by ER80S-G/MF6-55GP/ER80S-G show very low yield stress and low tensile strain value at failure. The average value of yield stress is close to the values of ER80S-G provided in the steel data sheet, but much lower than that of MF6-55GP [17]. The drastic reduction in the strength value can plausibly be explained by either the presence of manufacturing discontinuities and defects, such as lack of fusion or porosity, the coarsened grains, and/or the formation of brittle martensite-austenite (M–A) constituent within the bainite lamellar structures [18]. Further fracture mechanism will be discussed in the fractography section.

For each sample, there are two distinct ER80S-G/MF6-55GP interfaces, and the distance between the two interfaces; or the height of the deposited MF6-55GP, is approximately 22 mm. One interface is in the middle of the sample (clearly indicated by chemical etching), and the other is out of the gauge length and close to the fillet. As shown in figure 11, all three samples fractured in the MF6-55GP area, and the failures were all located in the gauge length. A little plastic deformation can be observed in the ER80S-G section, a roughly 2 mm elongation is detected from the scale.
Figure 12 shows the fracture surfaces after the tensile test was completed. The fracture surface is rather flat with a significant number of cleavage planes, indicating the occurrence of a relatively fast cleavage fracture due to the crack propagation. This fractograph shows a brittle-dominated quasi-cleavage failure, which is consistent with its relatively poor ductility. As mentioned before, the strength reduction can be rationalized into three reasons. Firstly, grain size is considered one of the most prominent microstructural characteristics that affects the mechanical properties. Although the deposited layer would be reheated by its subsequent layers and caused grain growth in the layer-by-layer deposition WAAM method, this process occurred in every layer except the uppermost one, which means the grain size increase is uniformly present. But further EBSD analysis on the grain size distribution is needed to confirm the deduction. Presence of M-A constituent is the second possibility for such a low tensile strength. As shown in figure 5 and figure 7, no M-A has been found in the metallurgy inspection. Another possible elucidation for the low yield stress and ductility of the ER80S-G/MF6-55GP/ER80S-G specimen is due to the microstructural discontinuities and defects induced during the additive manufacturing process. These defects can readily facilitate the brittle fracture by providing potential sites for cracks initiation and growth since they can act as strong stress concentration during tensile loading. This assumption was certified by the SEM images that a number of micro cracks were observed on different planes going toward the center of the fracture surface. The irregularity of the cracks can be correlated with the residual stress involved after deposition process [19]. As described in Arindam et al [20], crack extensions under quasi-static or tensile loading conditions often occur at very high ‘local’ stress intensities. It is well known that metallic parts fabricated using WAAM easily develop large stresses, which are due to the constrained thermal contraction during cooling between adjacent deposition layers. After the accumulation of several layers, the residual stress can reach a very high value. According to Shen et al [21], the residual stress in the normal direction (height direction) of the Cutoff (CO) walled component is significantly tensile, since it has the largest thermal gradient during metal solidification. The residual stresses can range from 60% to almost 100% of the materials’ yield
strength [22]. The large residual stress, combined with the tensile loading may account for such a low UTS (447.79 ± 24.32 MPa). As shown in figure 12, there are some cracks in the fracture surface, however, no crack has been observed in figures 5–7. This can be explained by the fact that the thin-walled sample for the microstructural inspection was cut from an object with only eight deposited layers, while the tensile sample had 60 deposited layers: the higher the number of deposition layers, the higher the residual stress in the as-built structure.

3.3. Wear properties

Three samples were cut from the as-built MF6-55GP structure and prepared for the wear test, with an average microhardness of 716.2 Hv. Commercial Cr12MoV die steel was also purchased and underwent a wear test under the same test conditions. The mean values of the weight loss under different loads of the MF6-55GP and Cr12MoV die steel were plotted in figure 13. The average material losses of the three MF6-55GP test samples under loads of 100, 200, and 300 N were 9.8 mg, 18.2 mg and 27.2 mg, respectively; and the values for the three Cr12MoV test samples were 7.1 mg, 16.5 mg and 27.9 mg respectively. Under a load of 100 N, the MF6-55GP weld metal had a much higher weight loss than the Cr12MoV die steel, but as the load increases, the weight loss difference between the two materials decreases. In other words, the MF6-55GP proved to be more stable when operating under high load, which indicates a better wear resistance than Cr12MoV. Figure 14 shows an OM micrograph of the worn surface of the MF6-55GP weld metal, which is characterized by long and parallel tracks. Deep plowing grooves, without wedge formation, appear on the worn surface; no clear delamination or wear debris was detected. This morphology indicates that micro-cutting is the predominant mechanism of wear [23]. As shown in figure 6, the MF6-55GP weld metal consists mainly of tempered or quenched martensite and this microstructure was proved to be resistant to deformation during wear process by Yi et al [24]. Therefore, it can be concluded that the MF6-55GP weld metal exhibits better wear resistance at higher loads than the Cr12MoV, under the conditions of a dry sliding wear test.

4. Conclusions

In this work, single-pass multi-layer and multi-pass multi-layer structures were deposited with ER80S-G and MF6-55GP wires using CMT-based WAAM. The goal of this study was to investigate the possibility of producing integrated wear-resistant components for severe wearing applications using WAAM instead of traditional processing methods. The microstructure and mechanical properties of the as-built objects were investigated; comparisons with die steel Cr12MoV were also made, and the results are summarized as follows:

(1) No apparent defects were found in the as-built single-pass eight-layer ER80S-G/MF6-55GP bimetal structure, including in the area surrounding the interface. Microstructural diversity between layers was
observed in both weld metals and was specifically obvious in the MF6-55GP portion. This can be explained by the different thermal histories experienced by the deposition layers.

(2) The hardness value gradually increased from the ER80S-G to MF6-55GP weld metal until reaching a peak hardness of 800 Hv at the final layer that underwent a quenching treatment. The hardness distribution in the MF6-55GP weld metal is highly related to the microstructural and morphological variation.

(3) The average UTS value of the ER80S-G/MF6-55GP/ER80S-G was $447.79 \pm 24.32$ MPa. The fractures always occurred in the MF6-55GP weld metal and exhibited a brittle-dominated quasi-cleavage fracture form. Such a low UTS may be the result of residual tensile stress superimposed with an external tensile load.

(4) As the wear load increases, the weight loss difference between MF6-55G and Cr12MoV decreases, this result exhibited that MF6-55G had a better wear resistance under a high wear load test condition. The microstructure of tempered or quenched martensite of MF6-55G weld metal may account for the resistance to deformation during wear process.

This study was present some preliminary results about producing wear resistant products by CMT-based WAAM process. The results are positive and affirmed the possibility of this present work. However, some issues are still not fully illuminated in this present work. Due to the similarity of the WAAM process to the conventional welding processes, the manufactured components are accompanied by residual stress and microstructural discontinuities and defects within the structures. More innovative strategies need to carry out on removing the residual stress and inherent defects. For instance, post heat treatment will be conducted on manufactured components in order to remove the stress and microstructural inhomogeneity, the changes of the microstructures and enhancement of the mechanical properties will be performed in the future work.

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