Tribology of a Fe–Cr–B-Based Alloy Coating Fabricated by a Controlled Short-Circuit MIG Welding Process

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Abstract The aim of this study was to study the microstructure and tribology of a Fe–Cr–B-based alloy coating deposited by a controlled short-circuit metal inert gas welding process onto a 1020 carbon steel substrate with varying input energies. Microstructure analysis showed that the as-deposited alloy consisted of $(\text{Cr,Fe})_2\text{B}$ particles embedded in a BCC solid solution matrix composed of Fe, Cr, Mn, and Si. The hardness of $(\text{Cr,Fe})_2\text{B}$ particles was 24 GPa. When the input energy increased during welding process, the deposition volume and dilution ratio were increased. As a result, $(\text{Cr,Fe})_2\text{B}$ particle volume fraction decreased from 44.6 to 37.2% and the bulk hardness decreased from 6.43 to 5.80 GPa. Dry sliding wear tests were carried out against a stainless steel counterface. The steady state coefficient of friction and the wear rate for the Fe–Cr–B-based alloy were independent of input energy. While the coefficient of friction for the Fe–Cr–B-based alloy was about 20% higher than for the 1030 carbon steel, the wear rate was about 90% lower. The dominant sliding wear mechanisms were adhesion and oxidation. Two-body abrasion wear test using alumina abrasives showed that the wear rate of the Fe–Cr–B-based alloy increased as the input energy increased and was about 90% lower than that of the 1030 carbon steel. The abrasive wear mechanism was microcutting.

Keywords Fe–Cr–B alloy · Metal matrix composite · CSC-MIG welding · Nanoindentation · Electron microscopy

Introduction

Dynamic machine components, such as pistons, shafts, journal bearings, cutting tools, pump parts, and aeronautical parts need to be protected against wear. Wear protection can be provided by metal matrix composites (MMCs), which are multiphase materials composed of hard reinforcing particles embedded in a metallic matrix [1]. One of these systems is the Fe–Cr–B-based alloys that contain hard boride particles, namely $(\text{Cr,Fe})_2\text{B}$, $(\text{Cr,Fe})_x\text{B}$, Cr_2B , $\text{Fe}_{1.1}\text{Cr}_{0.9}\text{B}_{0.9}$, and/or $\text{Cr}_{1.65}\text{Fe}_{0.35}\text{B}_{0.96}$, dispersed in a Fe-based solid solution matrix [2–15]. This system has been shown to exhibit excellent resistance against sliding and abrasive wear [6, 7]. The Fe–Cr–B-based alloys have commonly been applied as a coating using thermal spray processes, such as detonation gun [7, 14] and high velocity oxygen fuel (HVOF) [4–6, 11, 13]. Plasma-transferred arc (PTA) welding process has also been successfully used to fabricate a coating that provides a stronger metallurgical bond between the coating and the substrate and exhibits higher wear resistance when compared to thermal spray processes [11–13].

Welding techniques, such as gas metal arc welding (GMAW), plasma arc welding (PAW), and shielded metal arc welding (SMAW), are commonly used for hard-facing to protect the surfaces from wear, corrosion, and heat. The metal inert gas (MIG) process, known also as GMAW, is a welding process which uses an arc to melt a constantly fed electrode wire that is transferred to the substrate [16, pp. 19–22]. The controlled short-circuit metal inert gas (CSC-MIG) welding process is an improved version of the MIG welding process. In CSC-MIG process, the position and speed of the electrode wire are accurately controlled and thus leads to higher efficiency in material transfer, better stability, and lower heat input [17]. The CSC-MIG

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welding process was successfully used to deposit the overlay WC/Ni-based wire, which has a Ni sheath filled with FeSiB powder and WC particles, as high quality continuous beads without splatter [18]. The Fe–Cr–B-based alloy, which exhibits high wear resistance, has never been deposited by the CSC-MIG welding process.

Wear resistance of multi-phase materials, such as Fe–Cr–B-based and WC-based, depends on their wear mechanism, which is different from that of single-phase materials [19, 20]. Axen and Jacobson [19] proposed two modes of abrasive wear for two-phase materials, namely, equal wear rates of phases (EW) and equal pressure on phases (EP). In the EW mode, both phases are worn at the same linear rate and it corresponds to the upper limit of wear resistance (ideal state). In the EP mode, the matrix is worn independently, while the reinforcing phase is removed discretely. Thus, the reinforcing phase contributes slightly to wear resistance and the EP mode corresponds to the lower limit of wear resistance. The wear resistance of composite materials should fall somewhere between these limits [19] depending on a reinforcing particles size, matrix hardness, abrasive grit size, and inter-phase bonding [1]. It has been found that the volume fraction, shape, size, and orientation of the hard boride particles affect the wear performance of the Fe–Cr–B alloy [10, 12, 13, 15]. Welding the alloy onto a substrate, such as carbon steel, would cause dilution that changes the composition, microstructure, mechanical properties, and consequently wear performance.

The purpose of this work was to study the microstructure and tribology of the Fe–Cr–B-based alloy system deposited by the CSC-MIG welding with varying input energies. Microstructure characterization was carried out to identify the phases and their distributions, size, shapes, and volume fractions. Nanoindentation and microindentation techniques were used to measure the particles and the bulk hardness, respectively. Sliding wear and abrasive wear tests were carried out to study the friction and wear behavior in dry conditions. Post microstructural analysis was performed to investigate the sliding and abrasive wear mechanisms.

Experimental Procedures

Material

A commercial Fe–Cr–B-based alloy, designated as Armacor M, was selected for this research. It was received as a cored wire designed for twin wire arc spray (TWAS) process. The chemical composition of this alloy, as given by the manufacturer's data sheet, was 26.5–31.5% Cr, 3.4–4.2% B, 1.1–2.1% Si, and 1.1–2.2% Mn (all in wt%) with the balance being Fe.

CSC-MIG Welding Process

Single bead and multi-pass coatings (adjacent parallel beads) were deposited onto plain carbon steel (AISI 1020) plates, $100 \times 40 \times 3$ mm³ in dimensions, using the CSC-MIG welding system. This welding system was described in detail elsewhere [18]. The shielding gas was argon with a flow rate of 0.0189 m³/s (40 cfm). Table 1 lists the CSC-MIG welding process parameters used to fabricate three specimens designated as M1, M2, and M3.

Phase and Microstructure Analysis

For cross-sectional analysis, the specimens were cut perpendicular to the welding direction. For surface analysis, the multi-pass deposition beads were ground from the top surface to create a flat surface. For microstructure examination, mechanical properties measurements and wear tests, the specimens were cut, mounted, and then prepared, using standard metallographic techniques. They were ground progressively down to 800 grit using SiC papers followed by polishing using diamond solutions to 1 µm finish. After the final polishing with colloidal silica (0.05 µm), the specimens were cleaned in an ultrasonic bath filled with acetone for about 10 min to remove the residual debris and then dried in air.

Phase analysis was performed by x-ray diffraction (XRD) using Cu K α ($\lambda = 0.15406$ nm) radiation. XRD scans were carried out under the operation conditions of 40 kV and 20 mA with a 2θ step size of 0.01°/s and a dwell time per step of 1 s in the 2θ range of 30°–100°.

Microstructure and chemical analysis were carried out using a light microscope (LM), a field emission scanning electron microscope (FE-SEM), and energy dispersive spectroscopy (EDS). Secondary electron (SE) and back-scattered electron (BSE) micrographs were acquired at 15 kV accelerating voltage. Post-microstructural analysis of the worn surfaces of both sliding and abrasive wear was also accomplished using the FE-SEM and EDS to investigate the wear mechanisms. Quantitative analysis of the size and morphology of the phases was performed using commercially available image analysis software. Twenty BSE

Table 1 The CSC-MIG welding process parameters

Specimen	M1	M2	M3
Input energy (J/mm)	78 ± 5	220 ± 5	450 ± 10
Arc current (A)	100	200	300
Arc time (s)	0.01	0.01	0.01
Short-circuit current (A)	75	150	200
Short-circuit time (s)	0.01	0.01	0.01
Wire feed speed (mm/s)	183	217	250

micrographs of each specimen were employed to measure the volume fractions and aspect ratios of the phases. All average values in this research were compared by two-sided *t* student statistical tests using confidence interval of 99%.

Hardness and Elastic Modulus Measurements

Nanoindentation was used to measure the hardness H and reduced elastic modulus E_r of the phases within the specimens. The measurements were conducted at room temperature ($\sim 21^\circ\text{C}$) using a diamond Berkovich tip. The system is equipped with a piezoelectric scanner used to acquire topographical images similar to atomic force microscopy (AFM). The load-controlled indentation cycle was composed of loading for 5 s, holding at maximum load for 5 s, and unloading for 5 s. For each phase, 20 indents were performed to obtain mean values and standard deviations. The H and E_r were calculated from the load–displacement curves using Oliver and Pharr method [21]. Because pile-up was observed for all indents, post scanning of each indent was performed to measure the increased contact depth which is used to recalculate the actual contact area and correct the hardness and reduced modulus accordingly, as described [9].

Microindentation was used to measure the bulk hardness of the specimens using the Vickers microhardness tester at room temperature ($\sim 21^\circ\text{C}$). The indents were performed on a polished surface under a load of 500 g (4.91 N) with 40 indents onto specimens of each welding condition in Table 1 to determine the mean value and standard deviation.

Tribology Tests

Tribological properties and the behavior were evaluated by dry sliding and abrasion wear tests. Dry sliding wear tests were conducted using a custom-built linear reciprocating tribometer (ball-on-flat configuration) [22] in ambient air at room temperature ($22.6 \pm 0.5^\circ\text{C}$) and humidity ($19 \pm 5\%$ RH). The sliding wear tests were performed for two specimens of each welding condition in Table 1. Three tests were carried out on each specimen giving a total of six runs for each welding condition. The counterface was a full hard temper wear resistant stainless steel ball (type 440C) with a diameter of 12.7 mm and a hardness of 7.2 ± 0.2 GPa. The track length was 2 mm and a constant sliding speed of 3 mm/s was achieved for at least 95% of the track length. The total sliding distance was 20 m (5,000 cycles) and the reciprocating frequency was 0.632 cycle/s. The normal applied load was 85 g (0.834 N). The frictional force was recorded during the test to calculate the coefficient of friction ($\text{COF} = \text{frictional force}/\text{normal load}$) [23, pp. 209]. Depth profiles of the wear tracks were measured (three measurements per track, one at the middle and the others between the middle and the

ends) using a stylus profilometer. The wear volume was calculated by multiplying the average of the three measurements for the wear scar cross-sectional area by the track length (2 mm). Wear rates (in mm^3/Nm) of the specimen was calculated using Eq. 1:

$$\text{Wear rate} = \frac{\text{wear volume}(\text{mm}^3)}{\text{load (N)} * \text{sliding distance}(\text{m})} \quad (1)$$

For comparison and reference purpose, similar wear testing condition was performed on a medium carbon steel (AISI 1030) specimen. The hardness of this carbon steel was measured using the Vickers microhardness as 169 ± 3 HV (1.65 ± 0.03 GPa).

Two-body abrasion wear tests were conducted in ambient air at room temperature ($20 \pm 1^\circ\text{C}$) and humidity ($48 \pm 7\%$ RH) using a Loop Abrasion Tester designed for wear tests described in ASTM G174 standard [24] procedure (C). The specimen size of about $32 \times 8 \times 4$ mm was cut from the multi-pass weldment and ground from the top side of the tested surface down to a roughness less than $0.2 \mu\text{m Ra}$. The abrasion wear test was repeated three times for each specimen. The normal load was 100 g (0.981 N) and the spindle speed was 100 rpm. The test duration was 75 belt passes with a belt length of 1,295 mm; thus, the total abrasion distance was 97.1 m. The abrasive used was a $30 \mu\text{m}$ aluminum oxide microfinishing tape. The wear scars were measured at the middle using a non-contact laser profilometer. The wear volume was calculated using geometry calculation as described in ASTM G174 [24]. Abrasion wear rate in (mm^3/m) was calculated by dividing the wear volume by the total abrasion distance (97.1 m). For comparison and reference purpose, same wear test condition was performed onto a medium carbon steel (AISI 1030) specimen of the same reference material described above for sliding wear testing. It should be noted that for all repeated measurements, the average and standard deviation were calculated, and the result was reported in this article as (average $\pm$ standard deviation).

Results and Discussion

Microstructure and Phase Analysis

Figure 1 shows cross-sectional light micrographs of the Fe–Cr–B alloy deposited using CSC-MIG welding process with different heat inputs. Heat input (the amount of energy delivered per unit of length) is a function of voltage, current, and welding velocity [25]. When the voltage and current increased (Table 1) during the welding process, the heat input increased which caused higher melting masses of the electrode tip and substrate leading to higher

Fig. 1 Cross-sectional light micrographs of the Fe–Cr–B-based alloy deposited by CSC-MIG. (a) M1, (b) M2, and (c) M3

Fig. 2 SEM (BSE) micrographs of (a, b) M1, (c, d) M2, and (e, f) M3 specimens

deposition volume and weld dilution, respectively. While porosity was observed in M1 specimen, no porosity was observed at higher heat input within M2 and M3 specimens. There was no evidence of weld dilution in M1 specimen. However, the dilution for M2 specimen was 3 ± 2.4 vol% and that of M3 specimen was 19 ± 6.0 vol%, as measured using image analysis based on four micrographs for each welding condition.

Figure 2 shows SEM (BSE) micrographs of M1, M2, and M3 specimens. The as-deposited alloy consisted of two distinct phases, dark-contrast primary particles (β) and secondary particles (β') impeded in a light-contrast matrix. No cavities or cracks presented at the interface between the particles and the matrix. The distribution of the β particles

was uniform throughout the deposition (Fig. 1), while the β' particles distribution between the β particles were not (Fig. 2). As the heat input increases, the size and the distance between the β and β' particles slightly increased. Table 2 lists the quantitative data analysis based on these SEM (BSE) micrographs. The particle's (β and β') volume fractions for M1 and M2 specimens were equivalent but were lower for M3 specimen. However, the aspect ratios of β' particles of the specimens were equivalent.

The XRD results (Fig. 3) for the three deposited specimens show one series of peaks that correspond to a body centered cubic (bcc) α -Fe phase (JCPDS#006-0696). The second set corresponds to the orthorhombic structure boride phases, namely $\text{Fe}_{1.1}\text{Cr}_{0.9}\text{B}_{0.9}$ (JCPDS#072-1073) and

Table 2 Quantitative data analysis for M1, M2, and M3 specimens

Sample	Volume fraction (%)		Aspect ratio of the β' particles
	Particles (β and β')	Matrix	
M1	44.6 $\pm$ 3.9	55.5 $\pm$ 3.9	10.2 $\pm$ 3.3
M2	44.2 $\pm$ 2.7	55.8 $\pm$ 2.8	10.1 $\pm$ 4.2
M3	37.2 $\pm$ 2.6	62.9 $\pm$ 2.7	8.2 $\pm$ 3.7

Fig. 3 XRD results of the deposited alloy specimens

Cr_2B (JCPDS#003-4131). EDS analysis (Fig. 4a) shows that the β and β' particles (in Fig. 2) were rich in B, Cr, and Fe, while the matrix was rich in Fe, Cr, Mn, and Si (Fig. 4b). It should be noted that the EDS spectrums of the particles (Fig. 4a) were similar for all the three specimens (M1, M2, and M3) and within the same specimen as well. Likewise, the EDS spectrums of the matrix (Fig. 4b) were also similar for all specimens and within the same deposition. It is hence concluded that the dark-contrast particles (β and β') correspond to boride of Cr and Fe phase of the type M_2B and the light-contrast matrix corresponds to a BCC solid solution of Fe, Cr, Mn, and Si. For similar Fe–Cr–B-based alloys, it has been reported the boride phase particles as $(\text{Cr,Fe})_2\text{B}$ [2, 3]. The formation of the BCC solid solution phase is also consistent with the reported results [2, 5, 6, 9–11, 14].

Nanoindentation was used to measure the hardness and the reduced modulus of both the β particles as well as the mixed region, which involves the β' particles and the

Fig. 4 Typical EDS analysis of (a) the particles and (b) the matrix for M2 specimen

matrix. The nanoindentation tests were conducted at the middle regions of the specimens' cross-sections (Fig. 1). While the mechanical properties of the hard particle can be affected by the matrix properties, simulations by Yan et al. [26, 27] showed that the Oliver–Pharr method [21] can still be applied to measure the elastic modulus and hardness of the particle with sufficient accuracy if the indentation depth is within the particle-dominated depth. For our experiments, the particle-dominated indentation depth was estimated as explained in [27] and the indentation depth into the β particles was found sufficiently shallow to satisfy the condition set out by Yan et al. [26] for use of the Oliver–Pharr method [21] on particles in composites. Table 3 lists the hardness and the reduced modulus of the $(\text{Cr,Fe})_2\text{B}$ β particles and the mixed region for M1, M2, and M3 specimens. The hardness and the reduced modulus were each similar among the specimens. The hardness of the $(\text{Cr,Fe})_2\text{B}$ particles (β) exhibited a high value of about 24 GPa which was similar to the hardness of the boride layer containing Fe_2B , FeB , Cr_2B , CrB , and MnB that can reach up to 2,500 HV (~ 24.5 GPa) as reported by Ozbek et al. [28]. The hardness of the mixed region was in the range of 8–10 GPa. Figure 5 shows the typical load vs. depth curves for the indentation onto the particle and the mixed phase.

Table 3 Nanoindentation and Vickers microhardness results for M1, M2, and M3 specimens

Specimen	Nanoindentation				Vickers microhardness	
	β Particles		Mixed phase		(HV)	(GPa)
	H (GPa)	E_r (GPa)	H (GPa)	E_r (GPa)		
M1	23.6 ± 1.5	300 ± 24	8.1 ± 1.4	224 ± 15	655 ± 41	6.43 ± 0.41
M2	24.1 ± 1.0	318 ± 14	8.5 ± 1.6	226 ± 13	650 ± 58	6.37 ± 0.57
M3	24.5 ± 0.7	318 ± 10	9.8 ± 2.7	239 ± 19	592 ± 40	5.80 ± 0.40

Fig. 5 Typical load vs. depth curves for the indentation onto the particle and the mixed region for M2 specimen

The bulk hardness was measured using the Vickers microhardness under 500 g (4.91 N) load and the results are presented in Table 3. While the bulk hardness of M1 and M2 specimens were similar, that of M3 specimen was statistically lower. The bulk hardness of M3 specimen was decreased because of the weld dilution that reduced the B concentration and hence reduced the boride particles volume fraction (Table 2). This finding is consistent with previous studies [2, 3, 29], which have shown that as the B content increases, the fraction of the hard boride phase increases and consequently the bulk hardness increases. It should be noted that the hardness of the boride particles did not change with the heat input variation.

Sliding Wear

Dry sliding wear test of the Fe–Cr–B-based alloy deposited using CSC-MIG welding and the 1030 carbon steel was carried out against stainless steel counterface for 20 m sliding distance. Figure 6 shows the COF evolution as a function of cycle number and sliding time for M1, M2, and M3 specimens. The COF can be divided into two distinct regions as “running in” and “steady state”. The “running

Fig. 6 The coefficient of friction vs. cycle number and sliding time for the specimens when slid against stainless steel up to 20 m

in” period represents the first 200 cycles (315 s), as shown in small plot in Fig. 6, in which the COF started at about 0.1 and increased progressively until it became stable. The steady state COF was estimated by calculating the average and standard deviation of all the data points (Fig. 6) for the repeated tests for each specimen after excluding the first 200 cycles (“running in” period) and the results are presented in Table 4. The steady state COFs for M1, M2, and M3 specimens were equivalent and were about 20% higher than the carbon steel one.

The wear volumes of M1, M2, M3, and 1030 carbon steel specimens are listed in Table 4. The wear volumes of the M1, M2, and M3 specimens were equivalent. Similarly, the wear rates were equivalent, as shown in Table 4. The two distinct values of wear volume and wear rate for M1 specimen correspond to data from test-to-test variations that were not correlated to difference between the two specimens of this weld condition. That is, each reported average has data from each of the two weld specimens for M1. While the welding heat input variation affected the microstructure and the hardness, it was not significant

Table 4 Dry sliding wear results of the specimens slid against stainless steel counterface for 20 m sliding distance

Specimen	Steady state COF	Specimen wear volume ($\times 10^{-5} \text{ mm}^3$)	Specimen wear rate ($\times 10^{-6} \text{ mm}^3/\text{Nm}$)	Counterface wear rate ($\times 10^{-6} \text{ mm}^3/\text{Nm}$)
M1	0.887 ± 0.033	2.92 ± 0.44 3.27 ± 0.41	1.82 ± 0.28 2.04 ± 0.25	4.1 ± 0.5 53.1 ± 18.8
M2	0.889 ± 0.056	3.51 ± 0.73	2.19 ± 0.46	46.4 ± 17.2
M3	0.885 ± 0.027	3.57 ± 0.66	2.23 ± 0.41	15.9 ± 5.4
1030 Carbon steel	0.739 ± 0.044	34.80 ± 5.01	22.03 ± 4.43	13.7 ± 13.0

Fig. 7 Light micrograph of the wear scar on the stainless steel counterface slid against M2 specimen for 20 m

enough to change the sliding wear resistance. The wear volumes of the Fe–Cr–B-based alloy were about 90% lower than that of the carbon steel.

The stainless steel counterface was also worn during sliding, as shown in Fig. 7, and the wear rates of the counterface are presented in Table 4. For M2 and M3 specimens, the counterface wear rates were similar from test-to-test and a single average counterface wear rate was calculated for each specimen. However, the counterface wear rate for M1 specimens did not converge to a single average value and instead these specimens exhibited two distinct counterface wear rates. One wear rate ($53.1 \pm 18.8 \times 10^{-6} \text{ mm}^3/\text{Nm}$) was similar to that measured for M2 specimen, but the other one was significantly lower ($4.1 \pm 0.5 \times 10^{-6} \text{ mm}^3/\text{Nm}$) and more closer to the counterface wear rate measured for M3 specimen. As the boride particles volume fraction (Table 2) and bulk hardness (Table 3) of M3 specimen were lower than that of M1 and M2 specimens, it would be expected that the M1 and M2 specimens abrade the counterface at a similar rate, while M3 specimen would abrade it at lower rate. Thus, all of the measurements of the counterface wear rate agree

Fig. 8 (a) SE and (b) BSE micrographs of the wear track of M2 specimen after the wear test

with the hardness and changes in boride volume fraction except for the set of measurements for M1 specimen where the counterface wear rate was low. We hypothesize that this set of data may reflect some microstructural variation in M1 specimen that we have not rigorously studied yet. It is worth mentioning that M1 specimens were the most difficult to fabricate and while care was taken to prepare the specimens for three welding conditions used for this study in the same manner, some small differences in the microstructure, especially the boride distribution at the location where the wear test was conducted could explain these differences.

Fig. 9 SE and BSE micrographs of the wear track after sliding wear test for 20 m sliding distance for the specimens (a, b) M1, (c, d) M2, and (e, f) M3 with (g) a corresponding EDS analysis of the black regions in (d)

The sliding wear mechanism of the Fe–Cr–B-based alloy was investigated by post microstructure analysis, which showed similar characteristics of the worn surfaces for M1, M2, and M3 specimens. Figure 8 shows typical SE and BSE micrographs of the worn surface of M2 specimen after 20 m sliding distance. The entire worn surface was damaged. The damage included (Fig. 9) matrix removal, matrix oxidation, and boride particles (β) microcracking. EDS analysis (Fig. 9g) indicated that the matrix oxidation is oxides of Fe, Cr, Mn, and Si. It was observed that the oxides existed mainly in the regions where the matrix was removed. Moreover, Fig. 8 shows a build-up oxidized Fe-based material that could be a transferred material from the stainless steel counterpart. In addition, very few fine

grooves aligned with the sliding direction were observed (Fig. 8) due to the presence of hard particles (oxide debris and broken borides) that moved over the surface and scratched the surface.

The matrix elements oxidized due to the high temperatures at asperities (flash temperature), which could reach up to 1,000 °C depending on the applied load, COF, and sliding speed [30]. The oxide islands continued to grow and then were detached and pulled out by the adhesion and friction force (shear stress), which also caused material transfer from the softer surface (counterface) to the harder [30]. When the oxidized matrix was removed, the boride particles in contact were no longer supported and thus they were cracked, broken, and then removed. It can be

concluded that the main wear mechanisms of the Fe–Cr–B alloy, fabricated by CSC-MIG welding, that slid against stainless steel counterpart at room temperature were adhesive wear and tribo-oxidation (of the matrix) mechanisms, which occurred simultaneously. Similar wear mechanisms (adhesion and mild oxidation) were observed for dry sliding wear of the Fe–Cr–B-based alloy fabricated by spark plasma sintering (SPS) process with similar test conditions [9].

Abrasive Wear

Abrasive wear resistance and behavior were investigated by means of abrasive tester using alumina abrasives (with 30 μm particle size) for the specimens M2, M3, and 1030 carbon steel. The standard coupon size for abrasive wear test could not be fabricated using the M1 condition (Table 1) as the deposition volume was low and the multi-pass welding was not successful. Thus, M1 specimen was

Table 5 Abrasive wear test results for the specimens M2, M3, and 1030 carbon steel

Specimen	Wear volume (mm^3)	Wear rate ($\times 10^{-3} \text{ mm}^3/\text{m}$)
M2	0.16 ± 0.01	1.63
M3	0.22 ± 0.01	2.29
1030 Carbon steel	1.76 ± 0.03	18.20

not tested. Table 5 lists the wear volume of the specimens M2, M3, and 1030 carbon steel subjected to same conditions for the purpose of comparison. The wear volume of M2 specimen was lower than that of M3 specimen by about 29%. The wear volumes of M2 and M3 specimens were lower than that of the carbon steel by about 91 and 87%, respectively. Abrasive wear rates of these specimens, as shown in Table 5, followed same trend. The specimen M2 exhibited higher wear resistance because its bulk hardness was higher (Table 3) and therefore the penetration depth of the abrasive particles into the specimen surface was reduced [31–34]. Son et al. [3] and Yi et al. [29] also showed that the abrasive wear resistance increases when the bulk hardness and boride fraction increases. The results were also in agreement with previous studies [35, 36] that showed the increasing of wear resistance with the increasing volume fraction of the SiC and Al_2O_3 reinforcements. The hardness of the Fe–Cr–B alloy was much higher than that of carbon steel and thus the penetration depth was lower and the wear resistance was higher.

Abrasive wear mechanism was also investigated by post microstructural analysis of the worn surfaces. SE and BSE micrographs of the wear tracks of M2 and M3 specimens (Fig. 10) show continuous abrasive grooves on the surface. For both the specimens (M2 and M3), the groove widths were about 1–3 μm which was much smaller than the abrasive particles size (30 μm), indicating that the penetration depth of the abrasive was low. While no particle

Fig. 10 SE and BSE micrographs of the worn surfaces after abrasive wear test for the specimens (a, b) M2 and (c, d) M3

removal (pulled out) was observed, some microcracks in the primary boride particles (β) were detected, as shown in Fig. 10. Zum Gahr [37] showed that the abrasive wear occurs in four basic modes, which are microplowing, microcutting, microfatigue, and microcracking. While microplowing and microcutting are dominant on ductile materials, microcracking is more substantial on brittle materials [38]. The formation of the continuous grooves along minor microcracks within the boride particle (Fig. 10) implies that the predominant wear mechanism was abrasion in the form of microcutting mode.

Currently, the Fe–Cr–B-based alloy fabricated by the CSC-MIG welding process was tested at low loads, 0.834 N for sliding wear and 0.981 N for abrasive wear. The alloy showed high wear resistance as compared to the 1030 carbon steel. To understand better the wear performance for more severe applications, sliding and abrasive wear tests at higher loads need to be performed. Nevertheless, this investigation provided an indication of the potential for the CSC-MIG process to produce a high wear resistance Fe–Cr–B-based alloy.

Conclusion

The main questions addressed in this study concerned the investigation of microstructure and tribological behavior of the Fe–Cr–B-based alloy fabricated by the CSC-MIG welding process at varying conditions. The following conclusions can be drawn.

1. At low heat input, there was some porosity in the weldment and evidence that the bonding to the substrate was not complete. With higher heat inputs, the alloy was deposited and bonded metallurgically onto the 1020 carbon steel substrate with negligible cracks and porosity.
 2. The as-deposited coating consists of the hard $(\text{Cr,Fe})_2\text{B}$ primary and secondary particles embedded in the Fe-based BCC solid solution matrix composed of Fe, Cr, Mn, and Si. The hardness and reduced modulus of the primary boride particles are about 24 and 300 GPa, respectively.
 3. Welding parameters have affected the microstructure characteristics and the mechanical properties of this alloy. When the heat input increases because of the current and voltage increasing, the deposition volume and dilution increase. As a result, the $(\text{Cr,Fe})_2\text{B}$ particle fraction decreases and thus the bulk hardness decreased.
 4. Dry sliding wear test against stainless steel counterface shows that the COF and the wear rate of the deposited alloy were equivalent as a function of input energy.
- The independence of wear resistance on welding parameters variation can be considered as a good advantage for this welding process. The COF for the Fe–Cr–B-based alloy was about 20% higher than that of the 1030 carbon steel, whereas the wear rate was about 90% lower. The dominant sliding wear mechanisms were adhesion and oxidation.
5. Abrasive wear test using alumina abrasives revealed that the wear rate increased as the input energy increased. The wear rate of the Fe–Cr–B-based alloy was about 90% lower than that of the 1030 carbon steel. The abrasive wear mechanism was microcutting.

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