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Comparative study on wear behavior of as cast and spray cast hypereutectic Al-Si alloy

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Abstract:

The microstructure characteristics and wear behavior of a spray-deposited (SD) and the hot compressed Al-28Si alloys were investigated and the results were compared with corresponding SD and as-cast alloys. The microstructure, hardness and wear properties of the spray-deposited alloy was studied using X-ray diffraction, optical microscopy, scanning electron microscopy (SEM), microhardness measurement and wear tests. The microstructures of the SD alloy were composed of refined and uniformly distributed primary silicon (Si) and the formation a fine eutectic phase at the inter particle boundaries of an equiaxed grain morphology of the aluminum (Al) matrix. The decrease in the porosity level and further refinement of primary Si and eutectic Si phases in the hot pressed alloy. The wear tests indicate that the hot pressed spray formed alloy showed good wear resistance compared to both the SD and as-cast alloys. The improvement in the wear behavior is discussed in the light of microstructure evolution by the spray-deposition process. The worn surfaces were characterized by SEM-energy dispersive X-ray spectroscopy (EDS).

Keywords: Al-Si alloy, hardness, microstructure, spray deposition process, wear behaviour

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1 Introduction

Weight reduction technology has attracted increasing attention in the global automotive industry due to energy saving and environmental issues. The use of new materials with high strength and low density is the most effective method used for automobile weight reduction. Aluminum-silicon (Al-Si) alloys, representative examples of light metals, are commonly used in the automobile industry due to their enhanced properties. Hypereutectic Al-Si alloys find several applications in the aerospace, automobile and electrical industries owing to their low density, high wear resistance and high specific stiffness, as well as their thermal and dimensional stability [1]. By increasing the Si content, the Al-Si alloy exhibits excellent dimensional stability, hardness, low density and better tribological properties. Dry sliding wear behavior is more critically dependent on the amount and distribution of Si particles in the Al-matrix [2]. The relatively slow cooling rate associated with the casting process gives rise to the formation of a coarse primary Si phase and a high volume fraction of the needle-shaped eutectic phase in the Al-Si alloy results in a decrease in tension, ductility and increase in the wear rate. In general, fine, spherical and uniformly distributed primary Si particles in the Al matrix induce higher strength and better wear resistance [3], [4]. Spray forming is a new technology involving a rapid solidification process and is capable of producing a homogeneous fine equiaxed-grain microstructure [5]. The spray deposition process involves, in principal, the disintegration of the superheated molten metal using inert gas jets followed by its subsequent deposition the substrate. Briefly, as the liquid droplets fall, the gases in the atomizing chamber extract the heat from the droplets by convection, whereby the droplets cool and solidify. The primary function of the inert gas is to withdraw the heat from the liquid droplets, although the surface tension between the gas and liquid metal does play a part in breaking the liquid stream into droplets and preventing the deposit from the oxidation. Typically, an inert gas, He, N₂ or Ar, is used in the atomization process. The dynamic disintegration and deposition step enables this technique to produce advanced materials with enhanced mechanical and wear properties compared to conventional casting techniques. Further, spray forming allows the formation of fine metastable phases, increased concentration of alloying elements beyond solid solubility and the uniform distribution of second phase particles [6]. In the present investigation an Al-28Si alloy was fabricated via the spray

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deposition process and post-secondary process hot compression. The microstructural features, wear behavior of both the as-sprayed and hot pressed alloys were studied and compared with an as-cast alloy.

2 Experimental details

The chemical composition of the Al-28Si alloy is shown in Table 1. The alloy was melted in a graphite crucible using an electric resistance furnace and the casting process was done using the conventional green sand mold casting technique. The details of the spray forming set up employed in the present study have been described elsewhere [7]. In brief, the spray forming process employed an annular convergent-divergent free fall nozzle to create a spray of melt. In each run, 2.5 kg of alloy was melted to a super heat temperature of 100°C in a resistance heating furnace. The molten metal was passed through a ceramic delivery tube of 4-mm, the atomization gas comes in contact with the liquid stream in an atomization zone, which is located at 15 mm from the tip of the delivery tube and atomized by N₂ gas. The resulting spray was deposited on a Cu substrate resulting in a near net shaped preform. The details of processing variables are given in Table 2. The samples of square shape of 20 mm and length 50 mm were machined from the center of the preform. These were hot compressed after soaking them at 480°C for 60 min at a load of 50 kN. The samples were extracted from the as-cast (AC), spray-deposited (SD) and SD and hot pressed (SD + HC) for microstructural examination and wear testing. The porosity/density measurements were carried out on cylinder samples of diameter 10 mm and 50 mm length prepared from the center portion of the preform as well as on spray formed samples after hot compressing using a water displacement method based on the Archimedes' principle as per ASTM: B 962-08. The samples for microstructural examination were prepared by polishing using the standard metallographic techniques of grinding with Emery Paper of 1/0 (coarser), 2/0, 3/0 and 4/0 (fine) grades of successive fineness. The final polishing was done on a wheel cloth using Brasso and kerosene. The polished samples were etched with Keller's reagent. The microstructures of the samples were examined under a ZEISS Optical Microscope and using scanning electron microscopy (SEM) (Model: S-3400 N Hitachi Model). The hardness of the metallographically polished sample measurements was carried out using a Vickers Hardness Tester (Mattoon ATK-600) at an applied load of 300 g and a dwell time of 10 s. Micro hardness was the average of 10 measurements. The wear test was done under dry sliding conditions on a pin-on-disc type wear testing machine (Model: TR-20, DUCOM, as per ASTM: G99-05) having the counterface of a hardened EN-32 steel disc with hardness of 65 Rockwell hardness scale (HRC). The size of the pins for testing was 8 mm diameter and 30 mm length. The flat surfaces of both the pins and disc were polished to a surface finish of 0.3 µm before each test. Dry slidings were conducted under normal loads of 10, 20, 30, 40 and 100 N at a constant sliding distance of 5000 m for sliding velocities varying from 0.5 to 1.5 ms⁻¹. Each test was repeated 3 times with identical new samples on the fresh disk surface and the weight loss determined as a function of distance was used for the analysis of the wear rate. The hardness and wear resistance of as-cast, spray formed and hot compressed alloys are discussed in the light of its microstructural features and the morphology of the worn-out surfaces.

Table 1: Chemical compositions (wt %) of Al-28Si alloy.

Si	Cu	Fe	Ni	Al
28	0.056	0.025	00.035	Bal

Table 2: Primary processing parameters.

Melt super heat temperature (°K)	100
Atomizing gas	N ₂
Gas pressure (kg/cm ²)	4.5
Type of nozzle	Free-fall
Flight distance (mm)	390
Melt flow rate (kg/min)	2.48
Melt delivery tube diameter (mm)	4.00

3 Results and discussion

Figure 1 represents the X-ray diffraction (XRD) analysis of the AC and SD+HC alloys. It indicates that the phase constituents in the both alloys are Al rich phase (α -Al) and diamond cubic Si equilibrium phases. The diffraction peaks correspond to the reflections of atomic planes belonging to Al, Si and no reflections of crystalline planes of any intermetallic phase. The alloy exhibits high intensity peaks of α -Al and low intensity peaks of the Si phase. This indicates the overall equilibrium dissolution of Si in the Al lattice due to high under-cooling. No new phases were produced during the hot compression processing (Figure 1B). The optical micrograph of the as-cast alloy is shown in Figure 2A. It is composed of a primary Si phase exhibited in the form of coarse plate-like morphology with the size varying from 75 to 400 μm and an acicular form eutectic Si phase distributed randomly in the dendrite Al-matrix. SEM/ energy dispersive X-ray spectroscopy (EDS) microstructure of the as-cast alloy as shown in Figure 2B. The microstructure of the as-cast alloy consists of a dendritic structure of α -Al, a primary plate-like Si phase and needle and Fishbone morphologies of the eutectic Si phase. The EDS analysis illustrated in Figure 2B reveals that the gray contrast plates contain Si as a major element and chemical composition of different phases of as-cast alloy is shown in Table 3. Figure 3A shows the optical micrograph of the SD alloy. The microstructure of the as-sprayed alloy constitutes a homogeneous distribution of fine primary silicon particles ($\sim 5 \mu\text{m}$) eutectic silicon phase and the presence of fine, randomly distributed spherical shape porosity as high over 10.2%. The SEM micrograph of SD alloy is shown in Figure 3B. The micrograph clearly shows the homogeneous, globular and uniform distribution of primary Si particles in the Al matrix and the increased solid solubility of Si in the eutectic phase (point 1, Figure 3B) is high compared to the as-cast alloy. The phase compositions of the matrix phase, fine Si primary, eutectic phases with gray contrast, were determined using EDS as illustrated in the Table 4. The formation of porosity in the SD alloy perform, due to either gas entrapment or solidification shrinkage during the deposition process. In spray forming, the condition employing a less nozzle-to-substrate (flight) distance gives rise to an excessive liquid fraction on the deposition surface. Consequently, high gas velocity causes whipping of the liquid and entrapment of gas leading to the formation of large size cavities. The irregular morphology of pores observed in the top region of the preform supports this mechanism of porosity formation. Dissolved gas in the molten metal may also cause gas porosity as it would have the tendency to nucleate, grow, coalesce, and escape during solidification if there is a drastic reduction in gas solubility as the metal solidifies [8], [9]. Hydrogen is the only gas that is appreciably soluble in Al and its alloys, as described by Davis [10]. Its solubility varies directly with temperature and the square root of pressure. During the cooling and solidification of molten Al, dissolved H_2 in excess of the extremely low solid solubility may precipitate in the molecular form, resulting in the formation of gas pores. However, this kind of gas porosity has not been identified in the spray formed Al-Si alloys. It is thought that the dissolved H_2 may precipitate and escape into the vacuum melting and the protective N_2 atmosphere. Additionally, H_2 may also be released in N_2 atomization as the large specific surface area of the small droplets provides this opportunity. Moreover, solidification shrinkage can lead to the formation of pores if the temperature gradient in the mushy layer is low and interdendritic feeding is impossible. Cai and Lavernia [11] suggested that solidification shrinkage is significant only when too much liquid is present in the spray upon impingement [12], [13], [14]. The optical and SEM microstructures of the SD+HC alloy are shown in the Figure 4. The Al-matrix grains are equiaxed, and have a fairly wide size distribution ranging from 2 to 15 μm . The indication of fragmentation (Figure 4B), uniformly distributed Si phase whose size is a wide range between 2 and 10 μm and increased the volume fraction of fine Si particles subsequent to the hot compression processing. The roundness of the primary Si and the eutectic silicon phase can be ascribed due to the partial dissolution of these phases during hot compression as well as the reduction of the volume fraction of porosity. An equiaxed, homogeneous, uniform distribution of the fine primary Si phase in the Al-matrix and modified eutectic phase in the SD alloy can be obtained due to abundant nucleation of the melt due to rapid cooling in the flight and on deposition. The semi solid droplets receive an impact, deformation and fracture due to inter particle collision in the spray during flight and also impinging on a substrate effectively breaks the Si particles, due to re-melting of the solid component of the spray at the top of the deposition surface of semi-molten liquid pool with turbulent fluid flow conditions due to high velocity droplets. This effect induces shearing action and fragmentation of dendrites into debris particles [15], [16], [17].

Figure 1: XRD analysis of the Al-28Si alloy (A) as-cast alloy and (B) SD+HC alloy.

Figure 2: Microstructure of as-cast alloy (A) Optical and (B) SEM/EDS.

Table 3: EDS analysis of as-cast alloy of Al-28Si alloy.

Location	Compositions	Phase	Al-K	Mg-K	Si-K	Fe-K
Point 1	$\text{Al}_{81}\text{Si}_{19}$	Al-Si eutectic	77.44	0.00	19.04	2.37
Point 2	Si_{99}	Primary Si	0.23	0.05	99.72	0.00
Point 3	$\text{Al}_{99}\text{Si}_1$	α -Al matrix	96.92	0.00	1.27	0.05

Figure 3: Microstructure of the SD alloy (A) Optical and (B) SEM/EDS.

Table 4: EDS analysis of spray deposited (SD) alloy.

Location	Compositions	Phase	Al-K	Mg-K	Si-K	Fe-K
Point 1	$Al_{16}Si_{84}$	Al-Si eutectic	15.24	0.13	83.27	0.12
Point 2	$Al_{99}Si_1$	α -Al matrix	97.78	0.11	0.87	0.05
Point 3	$Si_{98}Al_2$	Primary Si	2.25	0.00	97.23	–

Figure 4: Microstructure of the SD+HC alloy (A) optical and (B) SEM.

Figure 5: Variation of wear rate with load at sliding velocity of 1.5 m/s.

The hardness of the AC, SD and SD+HC alloys are as shown in the Table 5. The results reveal an increase in the hardness of the SD alloy is around 46% higher than the as-cast alloy and there is a significant increase in the hardness of the SD+HC alloy as compared to the SD alloy. The fine and uniform distribution of Si particles

in the Al-matrix and an increase in the volume fraction of hard Si phases will lead to a higher constraint in the localized deformation of the soft matrix by the applied indentation load. The Si particles impart an appreciable obstruction to plastic deformation caused by the indentation which was restricted by the fine eutectic phase. The increase in hardness in the SD+HC alloy can be attributed to microstructural refinement, fragmentation of primary Si particles and densification of the alloy by reducing the porosity level commonly achieved subsequent to the hot compression process. Figure 6 represents the variation of wear rate as a function applied load at a constant sliding speed of 1.5 m/s. It is clearly revealed that the as-cast alloy has the highest wear rate compared to the SD alloy in the entire applied load range and SD+HC shows high wear resistance compared to that of the SD alloy. The increase in wear resistance of the SD alloy compared to the as-cast alloy is due to the microstructural features of the alloy. The size and distribution of the primary Si phase in the Al-matrix have significant influence on the wear behavior. During wear tests, extensive fracture of Si particles occurred, and the coarse Si phases appeared to fracture more frequently than the fine Si particulates [18]. The amount of fracture in Si increases due to the presence of the plate-like coarse of the Si phase increase in shear stress owing to the applied load, and the matrix in the surrounding area of the Si phases deform initially and experience a large degree of deformation. The stress concentration resulting from localized deformation depends on the size of the Si phases and the resultant stress concentration will be more severe at the interface between the coarse Si phases and the matrix compared to between the fine Si particulates and the matrix. This implication is supported by the present findings, which show that coarse Si phases fractured readily, and the wear resistance of the as-cast alloy was decreased. The reduction in wear resistance of the SD alloy compared to the SD+HC alloy, as shown in Figure 4 can be explained based on the microstructural features of the SD+HC alloy. The hot pressing of SD alloy results in the decrease in porosity level, much finer the eutectic Si and primary Si phases are more uniformly distributed as a result of less solid solubility in the Al-matrix. Fine and uniformly distributed particles (eutectic/primary) as a result of less solid solubility in the Al-matrix can be noted. Accordingly, the interfacial regions between the Si particles and the matrix become less and less prone to micro cracking. The micro cracking tendency becomes less predominant at low load. At high load, increasing the interface temperature between pin and disc limits the capability of the matrix to accommodate the Si particles and improves it due to better deformability of the matrix. Consequently, the micro cracking of the alloy is reduced. Under these circumstances, the fine primary Si facilitates more effective load transfer and thermal stability, leading to less wear response of the SD+HC alloy.

Table 5: Micro hardness of as-cast, SD and SD+HC Al-28Si alloy.

Sample	Micro hardness (VHN)
As-cast alloy	67 ± 7.03
SD alloy	97 ± 8.24
SD+HC alloy	116 ± 8.72

Figure 6: Variation of wear rate with sliding velocity at load of (A) 10 N and (B) 50 N.

The variation of wear rate with sliding speed at a load of 10 N is shown in Figure 6A. It is observed that at the low load, wear rate decreases with an increase in the sliding velocity and the as-cast alloy the wear rate is high as compared to that of the SD alloy. In addition, the SD+HC alloy shows less wear rate compared to AC and SD alloys. Figure 6B shows the wear rate of the as-cast SD and SD+HC alloys under different sliding velocities at a load of 50 N. It can be observed from Figure 6B that the wear rate follows a negative exponential pattern when the sliding velocity increases. This is due to the fact that at low sliding velocity more time is available for the formation and growth of micro welds, which increases the force required to shear off the micro welds to maintain the relative motion, due to which wear rate increases. As the velocity increases, there is less residential time for the growth of micro welds leading to less wear rate. With a further increase in the sliding speed (1.5 m/s) there is increase in the frictional force, frictional temperature, and the alloy becomes softer. Consequently intimate contact between the wearing surface and the rotating disc increases, results in a higher wear rate. Figure 7 shows the coefficient of friction (COF) as a function of the applied load. It can be seen that the COF decreases gradually with an increase in load for all the investigated alloys and the SD+HC alloy shows a low COF as compared to the AC and SD alloys. The value of COF at a higher load is about one half of its value at a low load in the load range investigated. It is suspected that at a low load, more time is available for the formation and growth of asperity junctions in the contact regions, which increases the force required to shear off the junction to maintain the relative motion, hence the high coefficient of friction.

Figure 7: Coefficient of friction as a function of sliding velocity at load of (A) 10 N and (B) 50 N.

3.1 Worn surfaces

SEM micrographs of wear surfaces of the as-cast, SD and SD+HC alloys under normal load of 10 N at sliding velocity of 1.5 m/s are shown in the Figure 8. The worn surfaces of an as-cast alloy (Figure 8A) reveals the presence of small craters, an oxidized surface and delamination by ploughing marks indicating a high wear rate and the spray formed alloy (Figure 8B) by mild scoring marks indicating a low wear rate. Whereas in Figure 8C the worn-out surface of the SD+HC alloy consists of only scoring marks indicating a low wear rate. This is due to the refinement of Si and eutectic phases in the SD+HC alloy. The SEM images of the worn surfaces of the as cast, SD and SD+HC alloys at a load of 50 N and sliding speed of 1.5 m/s are shown in Figure 9. The worn surface of the as-cast (Figure 9A) sample clearly indicates the mark of adhesive wear with a gross plastic flow of metal, metallic fracture of ridges, edge cracking and large dimples. The presence of hard primary Si and eutectic phases exists as coarse particles with a highly faceted nature in the cast alloy. Accordingly, it has relatively low bond strength with the matrix and the interfacial regions between the coarse Si phase and the matrix becoming very prone to micro cracking. Moreover, the sharp edges of the block-like primary Si particles introduce severe stress concentrations in the matrix of the alloy. These Si particles are more easily detached from the matrix and the debris of Si particles becomes well established in the matrix, resulting in deep and non-uniform grooves in the α -Al matrix. Typical surface damage, abrasive scoring marks, deeper and wider craters are clearly visible in the Figure 9A. Figure 9B shows the worn surface of the SD alloy. The smooth surface appears, small grooves with plastic deformation on the surface and oxidation of asperities, indicating the existence of mixed adhesive and abrasive wear. The low wear rate of the SD alloy compared to as-cast alloy is attributed to the microstructural refinement and large amount of fine and uniformly distributed Si particles in the Al-matrix. With increasing load, the fragmentation of Si particles and their ability to support the load results in the direct contact of matrix with the counterface and a large strain produced within the Al-matrix adjacent to contact surfaces. This might lead to the subsurface crack growth and delamination. The broken fine hard Si particles entrapped between the counterface and the alloy may act as third-party abrasers and be responsible for the fine longitudinal grooves on the worn surface. Worn surface of the SD+HC alloy as shown

in Figure 9C, reveals pronounced smooth sliding marks containing shallow grooves with smaller width, ridges and a few small dimples. The smaller width is due to the further refinement of the fine primary Si phase caused during the hot compression. In material with higher hardness, an abrasive wear tends to take place, while for material with lower hardness adhesive wear is encountered [4]. In the present study, for an SD+HC alloy with a Vickers' hardness (VHN) value of 126, abrasive wear has been found to be the main mechanism, while for the as-cast alloy with a VHN value of 69, adhesive wear has taken place. Thus, it can be implicit that the wear mechanism may also depend upon the hardness of the material.

Figure 8: SEM images of worn surfaces of (A) as-cast alloy, (B) SD alloy and (C) SD+HC alloy at load of 10 N and sliding velocity of 1.5 m/s.

Figure 9: SEM images of worn surfaces of (A) as-cast alloy, (B) SD alloy and (C) SD+HC alloy at load of 50 N and sliding velocity of 1.5 m/s.

4 Conclusions

The spray deposition process exhibits considerable microstructural refinement in the Al-28Si alloy. The primary Si and eutectic phases exhibit fine and uniform distribution in the Al-matrix in contrast to its coarse primary Si and dendritic morphology of eutectic in the as-cast alloy. Hot compression provides further refinement in the microstructure and reduced porosity of the preform. The wear resistance of the SD and hot compressed alloy is high compared to their counterparts of SD and as-cast alloys. The dominant wear mechanism for the SD alloy was the oxidative mechanism, and that for the as-cast alloy was an adhesive wear with a delaminating mechanism.

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