STRENGTH AND PLASTICITY

Comparative Study on H2O Steel Billets: Additive Manufacturing vs. Powder Metallurgy

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Abstract—Additive Manufacturing is one of the revolutionizing technologies of modern manufacturing industry. This technology has time and again proved that any material can be fabricated by it if the right parameters are employed. Other technologies such as powder metallurgy are also dominating the manufacturing industries owing to its potential flexibility in manufacturing complex shapes. In this work, an attempt has been made to fabricate cylindrical components through both the techniques, and their properties are compared. A study on metallurgical properties revealed that both provide similar microstructures, but powder metallurgy yielded better mechanical properties. It has been also observed that the tribological properties are better in additive manufactured components. The reason for this behavior has been studied and discussed.

Keywords: direct metal laser sintering, additive manufacturing, powder metallurgy, H2O steel

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INTRODUCTION

Additive manufacturing (AM) can be defined as a manufacturing technology in which parts or components are built through sequential addition of material in the form of layers through digital data. ASTM has mentioned AM as the process that joins materials to make desired products from a 3D CAD data, generally one layer upon another layer, in contrast to material removal technologies that are employed to fabricate the same articles [1].

After years of extensive research and development the AM is now extended to various fields like biomedical [2], automotive [3], aerospace [4], health care devices [5], jewelry, toys, and recreation [6]. The short product development cycle and high product quality [7] that emerges out of the AM technology, and its ability to fabricate functionally graded materials [8] have made it one of the most sought-for and well-established technologies in modern industries.

Before the AM technique, Powder Metallurgy (PM) came into existence giving high scope for production of complex geometries and micro-manufacturing. Sintering is one of the prominent techniques used for manufacturing of products through metallic and ceramics powders.

Figure 1 illustrates the main production steps for Metal AM and PM. For both processes, the powder is

used as a raw material. However, for Metal AM only spherical powders are applicable, since the flowability of the powder is a key parameter for successful manufacturing. For PM processes, powder flowability is not crucial. Hence, powders of irregular shape can be used.

From Fig. 1 it is clear that the main difference in the production route is the steps in manufacturing [9].

Both AM and PM processes require heat treatment, post-processing and finishing operations. Thus, it can be said that the main difference in the production chain is that the AM route enables the manufacturing of components with the freedom of design, along with complex geometry and high mechanical properties. That makes the AM a promising production technique.

The current study focuses on determining suitable manufacturing process especially in terms of improved mechanical properties such as wear, hardness, and better flexibility of process to produce complex parts with in a minimal duration. PM is an expression encompassing a wide spectrum of techniques where components are fabricated using micro- and nano-powders. To a large extent it eliminates the material loss with over 90% of material conversion rate thereby reducing the overall cost of the product. The powder used in PM process is compacted in a specifically designed mold, which is termed as green compact. It is

Fig. 1. Production chain steps in Metal Additive Manufacturing vs. Powder Metallurgy.

then sintered using a high temperature furnace with or without pressure applied to the part while heating. Basic steps involved in the PM are shown in Fig. 1.

The Direct Laser Metal Sintering (DMLS) method is one of the popular AM processes due to its ability to fabricate metal parts with high dimensional accuracy and its capacity to produce near net shape components. However, to establish its suitability to replace other manufacturing technologies, it is imperative that the properties of fabricated parts such as tribological behavior and hardness should be studied. The standard process parameters for SLM/DMLS process are Hatching Distance (HD); Laser Power (LP); Laser Scan Speed (LSS), etc.

Many tools are made of steel material and they are subjected to severe wear loading [10]. However, only limited sources are available which studied the effect of laser process variables on the tribological properties of laser sintered steel parts.

Ramesh et al. have manufactured parts using iron powder by the DMLS process [11]. These parts demonstrated improved density, microhardness, and radical enhancement in tensile strength, ductility, and wear at lower scan speed.

Keshavamurthy et al. have optimized the process variables of AM process for direct metal deposition of H13 tool steel based on Taguchi technique with L9 orthogonal array [12]. They optimized the process parameters such as: LP of 400W, LSS of 200 mm/min, and powder flow rate of 1 g/min.

Kazantseva et al. have studied SLM processing of stainless steel with magnetic properties [13]. It was found that the microstructure of SLM-made marag-

ing stainless steel differs from the conventionally produced one. This was further confirmed in [14].

Khorsand et al. assessed the effect of PM process variables on the fatigue strength and tribological characteristics of Fe-1.75, Ni-1.5, Cu-0.5, Mo-0.6 C sintered low alloy steel [15]. The results show that those properties were considerably increased when efforts such as heat treatment were undertaken to reduce the porosity.

Ceschini et al. also studied the effect of PM process variables on the tribological behavior of the Fe-C-Mo and the Fe-C-Cr sintered steel under dry and abrasive test conditions [16]. It was stated that Fe-C-Mo steel performed better than the Fe-C-Cr steel for the same process variables.

Senthur et al. carried out wear studies under dry conditions on the sintered and hot extruded material with the combinations such as Fe-1%C as base element alloyed with tungsten and titanium and a few other elements [17]. It was reported that an increase in the sintering load had improved the wear resistant property of the material under consideration.

From the vast literature review, it is found that considerable amount of work is carried out by researchers to evaluate the physical and tribological properties of parts produced by both processes. However, no information is available on the studies comparing them based on their physical and tribological properties.

The present work is aimed at comparing the physical and tribological properties of steel parts (EOS Steel H20) produced by PM and AM processes.

MATERIALS AND METHODS

The H2O powder of 20- μ m particulate size was used to fabricate the test specimens. The powder that was used had spherical shaped grains with a few agglomerations. The specimens were built layer-by-layer using a DMLS EOSINT M250 machine at Central Manufacturing Technology Institute, Bangalore. Using a dial indicator, the top surface of the steel building platform was adjusted to be parallel with recoating blade, maintaining an accuracy of 0.01 mm. The H2O powder was thoroughly rammed into the supplier tank so that all the trapped air was removed. The powder layer thickness was maintained at about 20 μ m. The beam had diameter of 0.40 mm to melt and fuse the powder. The LSS was varied between 50 and 125 mm/s, while the HD and hatch width were maintained constant at 0.3 and 5 mm, respectively.

Nitrogen was supplied to the chamber throughout the process to avoid oxidation and other types of contaminations to the specimens.

An up–down stripe pattern was employed for the laser scan during AM process.

The DMLS parameters are presented in Table 1. The LP was maintained at 240 W and the LSS will be optimized for producing sound specimens with good metallurgical properties.

Similar to AM, specimens were produced by PM process at METMECH ENGINEERS, Chennai. In the PM, ‘sintering’ was selected as a very feasible and widely used method for fabrication of PM related products. Sintering follows the compacting and forming procedure to fabricate a solid mass of material by application of heat and/or pressure without melting it to the point of liquation. Table 2 lists the details of the process parameters that were employed in the manufacturing of the specimens using PM.

The green compact was placed in an electric furnace at 950°C in nitrogen atmosphere to sinter the component. The steel powder is compacted with applied load of 220 MPa using a manual press. Sintering process applies pressure to the powder to get compacted (green part) in order to obtain strength and integrity. The temperature is generally maintained below the melting point of the major constituent of the material.

In both AM and PM processes the same Direct steel H2O powder was used as material for manufacturing of wear specimens, examination, and study. This material possesses certain characteristics like high strength, hardness, wear resistance, and surface density, which makes it suitable for tooling industries especially for direct tool applications such as injection molding inserts and dies for casting.

Table 3 shows various mechanical, physical, and chemical properties of H2O steel powder.

After fabricating the specimens through AM and PM technologies, certain surface morphological char-

Table 1. Built-up part DMLS process parameters

Sl. No.	Built parameters	Range
1	Hatching distance	0.3 mm
2	Hatch width	5 mm
3	Layer thickness	20 μ m
4	Specimens built size	$\varnothing 12 \times 8$ mm
5	Fabrication time	3 h
6	Laser power	240 W
7	Laser beam diameter	0.4 mm
8	Laser scan speed (LSS)	50–125 mm/s

Table 2. Process parameters of PM process

Sl. No.	Process parameters	Range
1	Temperature	900–950°C
2	Sintering time	1–4 h
3	Sintering atmosphere	Nitrogen
4	Specimens size	$\varnothing 20 \times 10$
5	Heating rate	5°C/min
7	Relaxation time in compaction	2 min
8	Fabrication time	4 h
9	Compaction time (4 specimens)	1 h

acteristics and microstructural properties were studied by optical and Scanning Electron Microscope (SEM) ZEISS JSM-6480 LV SEM. DE-WINTOR inverted trinocular metallurgical microscope was used to study the phase contrast and evaluate the grain size characteristics.

Micro Vickers hardness, using a microhardness tester, Clemex dual CMT 07-895 (Canada), was evaluated at a constant load of 0.3 g and with a dwell time of 10 s at different locations in built directions of the built-up parts.

Testing was performed according to the ASTM E384 requirements, with at least 8 probes for each specimen.

To derive the density of the sintered parts, first the samples were weighed in air and then in water and applying the Archimedes principle, the values were calculated. Density is measured using the standard procedure such as dividing the mass of a substance by its volume. Sartorius Electronic Balance (MC 210S) with the least count of 0.01 mg was used for the density measurement.

Pin-on-Disc Wear Tester (Ducom instruments: TR-20LE-PHM400-CHE600) instrument was used to measure the wear behavior of the DMLS and conventional sintering specimens. These experiments were conducted with a load of 20 N as the constant load applied for about 10 minutes at 803 rpm against a counter disc, which is made of EN31 steel. The

Table 3. Properties of H20 direct steel powder

No.	Properties	H20
1	Tensile strength (MPIF 10)	1100 MPa
2	Young's modulus	180 GPa
3	Hardness	350–420 HRC
4	Elongation at failure	4.0%
5	Composition after DMLS	Fe + Cr, Ni, Mo, Si, V, C
6	Density (skin area)	7.8 g/cm ³
7	Density (core area)	7.0 g/cm ³
8	Porosity	<0.5%
9	CTE	13–15 × 10 ⁻⁶ m/m K
10	Thermal conductivity	15–18 W/mK
11	Max operating temperature	1100–1470°C

dimension of the counter disc used in the wear study is 160 mm diameter with a thickness of about 8 mm having the 60 HRC hardness. A cylindrical and flattened sintered specimen of dimension 10 mm in diameter and height of 35 mm served as a pin. Before every friction and wear test, the surface roughness (RA) of sintered H20 specimen and counter disc on the wear tester were maintained at centerline average (CLA) values of 0.9 and 0.8 μm , respectively. Prior to every test, the specimens and disc were cleaned with acetone.

A pin on disc tribometer was used for wear studies. The sliding velocity was maintained at 1.26 m/s and load was maintained 20 N for both AM and PM processes. Volumetric wear rate was evaluated by measuring the weight loss during the sliding process.

Fig. 2. Optical micrographs of specimens at various LSS: (a) 50; (b) 75; (c) 100; (d) 125 mm/s.

RESULTS AND DISCUSSION

The results of metallurgical characterization of fabricated materials are as follows. Figure 2a shows the microstructure of the AM component. The LSS was maintained at 50 mm/s. The lower speed of laser produced good fusion of the powder particles. The lower speed and higher specific heat input resulted in conversion of steel powders to partially transformed martensite in non-transformed ferrite matrix. Fine dispersions of alloy carbides are also present in the matrix. The lower speed of the laser produced higher heat in conversion of martensite and the hardness is higher out of all the specimens, which is confirmed by hardness testing.

Figures 2b–2d illustrate the microstructure of the specimens for which the laser speed was maintained at 75, 100, 125 mm/s, respectively. As opposed to the LSS of laser which produced good fusion of the powder particles, the higher LSS and lower heat input in these three cases resulted in conversion of steel powders to partially lower transformed martensite in non-transformed ferrite matrix. Fine dispersions of alloy carbides are also present in the matrix. The corresponding hardness is also measured as lower value. Marginal reduction in grain size is observed between the different speeds of the laser heat.

Figure 3a shows fine pearlite in ferrite grains with some transformed martensite. The microstructure is more uniform and no large cavities are observed.

Figure 3b shows the microstructure as fine carbides and least transformed martensite. Some large islands of pores are observed in the matrix but fusion of the matrix has taken place.

In Fig. 3c one can see good fusion of the parent metal powder due to compacting and sintering, but lower transformation of martensite is observed. The

Fig. 3. Optical micrographs of conventionally sintered specimens for (a) 1; (b) 2; (c) 3; (d) 4 h.

grains are coarser. In addition, it can be seen that fewer cavities are present between the grains, and it indicates the effect of compacting.

It can be seen from Fig. 3d, that compacting and sintering result in good fusion of the parent metal powder, but lower transformation of martensite is observed. The microstructure is more uniform and no large cavities are observed. Samples No (c) and (d) have coarser microstructure and more pearlite grains in ferrite phase, which results in marginal increase in hardness.

The SEM micrographs of AM samples are shown in Fig. 4. The microstructure of the sample produced with LSS of 50 mm/s is shown in Fig. 4a. The magnification is $750\times$, which has resolved the microstructure which is tempered martensite with the presence of

secondary phases of alloy carbides which have formed during cooling [11, 18]. The grains are uniform and undergone fusion and transformation to martensite. No pores between the grains are observed in this field and the martensite is more resolved as matrix with secondary phases. Complete and effective fusion of the matrix is observed, and it caused higher homogeneity to give higher hardness.

Figure 4b shows the SEM of the sample made with higher LSS, namely, 75 mm/s. In this sample, the surface is more uneven and some large crests and troughs are present. The microstructure shows large and fine-grained matrix. The matrix shows no pores between the grains of the powder. The SEM magnification at $750\times$ resolved the secondary phase alloy carbides. The SEM micrograph shows the pulled-out particles which caused the troughs formed. High speed of the laser caused lower time for the particles to undergo sintering and transformations.

In the microstructure of the sample, manufactured at an LSS of 100 mm/s (see Fig. 4c), secondary phases are uniformly distributed in primary tempered martensite matrix that has formed by laser curing/melting process [18]. The secondary phases form lumps in certain locations. The un-sintered particles also show some pull outs from the matrix.

The LSS of 125 mm/s (see Fig. 4d) has also resulted in the secondary phases becoming uniformly distributed in primary tempered martensite matrix that has formed by laser healing/sintering process. However, deeper cavities are observed in this sample along the boundaries of the carbide phases/ grains. The un-sintered particles also show some pull outs from the matrix.

For the specimens sintered for 1 h, SEM micrograph (see Fig. 5a) shows fine-grained morphology of

Fig. 4. SEM images of the AM samples with the various LSS: (a) 50; (b) 75; (c) 100; (d) 125 mm/s.

Fig. 5. SEM images of PM samples after conventional sintering for (a) 1; (b) 2; (c) 3; (d) 4 h.

tempered martensite with fine pores in between the grains. Sintering in furnace has produced uniform microstructure. The grain boundaries are resolved and the pores present in the matrix are uniform. This might be due to insufficient compacting pressure.

The SEM micrograph of specimens sintered for 2 h (see Fig. 5b) shows distribution of the alloy carbide particles in fine tempered martensite matrix. More pores between the grains observed here are probably due to insufficient compacting pressure. This field shows a large area of rough surface caused by lower compacting pressure, and subsequent sintering failed to bond the particles. The primary and the secondary phase distributions are homogenous. The SEM also shows the fine homogenous microstructure of tempered martensite with the carbide particles distribution at $750\times$.

Figure 5c shows the SEM image of specimen that was sintered for 3 hours. It is seen that there is fine uniform martensite matrix with a distribution of carbide particles. The pores present in the matrix after sintering are fewer and finer.

Finally, the SEM image of the specimens sintered for 4 h (see Fig. 5d) shows fine homogenous grains of primary and secondary phases. The primary phase is tempered martensite and the secondary are the precipitated carbides. The field shows better fusion with fewer pores between the grains than in the other three images.

The microhardness of each sample is measured at 8 different locations on the surface at different distances from the center. Pronounced variations in the hardness from one location to another in each of these samples can be attributed to the presence of sporadic

porosity, which affects the hardness. However, the overall hardness of the samples can be assessed using the average of the 8 values measured for each sample. The variations of the average microhardness at different LSS along the built direction are shown in Fig. 6a. Significant dependence of microhardness of the built-up parts on the sintering speeds is observed, wherein, a decrease in LSS results in increased microhardness of the built-up parts.

A hardness of 370 HV is spotted on the specimens sintered at 50 mm/s. Lower laser (scan/sintering) speeds and the increase in the density of the sintered components resulted in the specific increase of microhardness. On the other hand, the porosity, voids, unmelt regions, and a few cracks found in the specimens were the reason for the poor microhardness at specific locations. At lower sintering speed, the presence of hard nitrides and oxides of iron acting as barriers for dislocation motion contributed to the increase in hardness of sintered parts [11, 19, 20]. Increased LSS resulted in lower microhardness and density.

The variation of local microhardness of the samples made by PM at different distances from the center of the sample is shown in Fig. 6b for the four samples that were sintered for different durations. As in the case of corresponding results for the AM samples, there are significant variations in microhardness from one location to another. It is also observed that there is radical disparity in microhardness of the built-up parts with change in sintering duration, wherein an increase in sintering hours results in increased microhardness of the fabricated parts. A peak microhardness of 331.2 HV is observed for built-up parts sintered for 3 h. The improvement in the microhardness of those parts

Fig. 6. Variations of microhardness (a) with LSS for AM samples; (b) with various sintering time for PM samples.

Fig. 7. Density variation: (a) for AM samples with LSS variation; (b) for PM samples with sintering time variation.

at increased sintering hours can be endorsed to the higher density of the finished product.

Figures 6a, 6b show higher microhardness of the specimens fabricated by AM process comparing to the PM-fabricated specimens. It is because of the higher sintering temperature up to 1200°C and also the sintering beam diameter (0.40 mm) used in the AM. There is a large amount of heat energy absorbed by specimens compared to conventional sintering (950°C) process. Thus, the higher hardness is obtained in specimens fabricated by the AM compared to those obtained by the PM.

The improvement of the microhardness of the built-up parts at lower LSS can be attributed to the superior density of the built-up parts.

Figure 7a shows variation in density of the AM parts with different LSS. At lower LSS, an enhance-

ment in the density as well as a decrease in porosity is noticed in the sintered parts. Lower LSS ensures higher energy absorption by the powder during the time of fusion, resulting in better melting and greater density. These interpretations are in agreement with those by several other researchers while sintering iron and steel powders [21]. Increased LSS results in lower energy absorption by the powder during the process, decreases density of the samples, and additionally increases porosity resulting in lower microhardness.

Figure 7b shows the dependence of density of the samples made by conventional PM on the sintering time. An enhancement in the density of PM parts was observed along with a decrease in its porosity for every increase in sintering time. Up to 3 hours of sintering time, the density of the sintered parts increased. For 4 h, the density seems to reduce because of a minor experimental error. The observed increasing of micro-

Fig. 8. Variation of wear rate with respect to time. The specimens built at an LSS of (a) 50, (b) 75, (c) 100, (d) 125 mm/s.

hardness at longer sintering times may be due to the improved melting and increased density.

Comparison of density of both AM and PM specimens shows the difference in density values of sintered parts having an LSS of 50, 75, 100, 125 mm/s, respectively. It has been found that there is an augmentation in the density and a drop in the porosity of the sintered parts, while a low sintering speed is maintained. Lower LSS enables higher energy absorption by the powder during the time of sintering, resulting in improved melting and higher density.

In the case of PM, different sintering span of time is used (1, 2, 3, 4 hours) for the improvement in the density and decrease in porosity of the sintered parts. The longer the sintering, the higher is the energy absorbed by the powder, resulting in improved melting and higher density of the part. A reduction in porosity is obtained by increasing the sintering temperature and, even more, the compacting pressure. These findings correspond to several other researches on sintering iron and steel powders by PM process [20].

The above graph shows the higher density obtained in the specimens fabricated by the AM process compared to the PM-fabricated specimens. It can be explained by a higher sintering temperature up to 1200°C. Besides, the sintering beam diameter of 0.40 mm used in the AM ensures a large amount of heat energy absorbed by specimens compared to conventional sintering (950°C) process, which results in the higher density of specimens obtained by AM compared to PM specimens.

Figure 8 shows the variation of wear rate of AM parts with respect to time and depending on the process parameters considered for wear tests, such as load 20 N, sliding velocity 1.2 m/s, sliding distance 756 m, and duration 10 min.

Figure 8 shows the results of wear testing for the four samples made by AM. The mean wear rate of each of the samples is calculated using the average wear rate values obtained at different times during the wear test for that sample. It has been found that wear properties degrade with respect to the decrease in LSS. This is

Fig. 9. Variation of wear rate with respect to sintering time: (a) 1; (b) 2; (c) 3; (d) 4 h.

attributed to the fact that the decrease in LSS resulted in improvement in hardness and strength of fabricated parts. The reduction in wear loss of the sintered parts can also be attributed to the inclusions such as hard oxides and nitrides in those final products [11].

The process parameters considered for wear test of PM specimens are load 20 N, sliding velocity 1.2 m/s, sliding distance 756 meters, and duration 10 minutes.

Figure 9 shows the deviation of wear rate of built-up parts at different sintering time. It is observed that the rate of wear decreases with increased sintering hours. This can be attributed to the fact that the increase in sintering results in enhanced hardness properties. Improvement in the hardness and density will lead to improved wear resistance. There are 'n' numbers of reasons for the increase and decrease in wear resistance.

Figure 10 shows the lower wear rate obtained by the specimens fabricated by AM process compared to PM-fabricated specimens, because decrease in LSS has caused the improvement in hardness and has made the parts stronger. Any improvement in hardness and strength will lead to superior wear resistant properties.

An increase in density and hardness does result in improvement of wear resistance of PM materials, and decrease in porosity and increase in sintering temperature result in lower wear rate.

Figures 11, 12 present macroimages of AM and PM specimens after wear test correspondingly. Figures 11 and 12 show the specimens (fabricated by AM and PM, respectively) subjected to wear tests, in which areas where the oxidizing (rust) starts due to the rise in temperature are clearly seen.

Figure 13a shows the relationship between the LSS and coefficient of friction for the parts built by AM and PM processes. It is to be noted that an increase in LSS has resulted in poor coefficient of friction. The reason for this is associated with the large amount of porosity in the as-prepared specimen. The reduction in the part density results in the reduction of roughness, thereby reducing the friction occurred because of the sliding of mating parts. Additionally, a decrease in laser (scan) speed will result in an increase in oxide and nitride contents in Fe, as mentioned in previous sections. This eventually results in the breakdown of inclusions during sliding, leading to greater exposure

Fig. 10. Variation of wear rate for AM and PM specimens.

Fig. 11. AM specimens subjected to wear test.

Fig. 12. PM specimens subjected to wear test.

Fig. 13. Variation of friction coefficient of AM and PM specimens.

of contact surfaces,—are mainly responsible for an increase in the coefficient of friction.

Figure 13b shows the effect of sintering hours on the coefficient of friction. It is observed that an increase of sintering time can cause the reduction of the coefficient of friction. Figures 13a, 13b show the lower coefficient of friction demonstrated by the specimens fabricated by AM process compared to PM-fabricated specimens. It is because the decrease in LSS has resulted in improved hardness and strength of laser built-up parts. An increase in the coefficient of friction is observed due to an improvement of hardness and strength of the specimens.

The AM specimens show drastic improvement in density, hardness, wear, and coefficient of friction compared to PM specimens due to the decreased LSS, when higher is the energy absorbed by the powder resulting in improved melting of particles.

CONCLUSIONS

(1) The specimens were successfully fabricated from H20 steel powder, by AM with varying laser scan speeds from 50 to 125 mm/s in periods of 25 mm/s and by PM with various sintering time from 1 to 4 hours in step of 1 hour.

(2) The optimum compacting pressure and sintering temperature for conventional sintering (PM) is fixed at 240 MPa and 950°C, whereas the best sintering time is 3 hours.

(3) The LSS has a great impact on the microstructure of the built-up parts, and the optimal microstructures can be produced when being further heat-treated to obtain required physical properties.

(4) The laser built-up parts sintered at a lower LSS have a higher density and microhardness. An increase in the coefficient of friction is observed for parts built at lower LSS of 50 mm/s compared to higher scan speeds.

(5) Increase in wear rate is observed with increased sintering hours in conventional sintering. An increase in coefficient of friction is observed for parts sintered at a time span of 1 hour. Increased density of the PM parts will result in higher probability of asperity interaction, which in turn can increase the friction during the sliding motion of the mating parts.

(6) The sintering time has intense influence on the microstructure of built-up parts. A higher density and microhardness are observed for 3 h sintering of a green compact part.

(7) SEM analysis revealed that there is a good bonding between the particles because of reduced LSS, higher heat energy absorbed by the powder during the process which results in improved melting and higher density. The improved strength could be expected due to enhanced density of the AM parts in agreement with [22].

(8) It is observed that wear and coefficient of friction are lower at the LSS of 50 mm/s, in AM parts when compared with the parts built through PM.

(9) Hence, it has been found that AM specimens show superior properties in terms of reduced wear rate, friction coefficient, increased density and microhardness with respect to parts built through PM.

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