

Investigation of the Effects of Solution Rapid Treatment on Wear Properties of AlSi10Mg Alloy

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Abstract

The purpose of this study was to evaluate the tribological response of laser powder bed-fused AlSi10Mg after an innovative rapid T6 heat treatment and to compare it with that of as-built and conventional T6 conditions. Four rapid T6 routes, each with a significantly shortened solutionization time (5–10 min at 510–540°C + aging), were compared with the standard T6 route (540°C, 2 h + aging). Microstructure (optical microscopy, SEM, and XRD), hardness, and dry sliding wear performance (ball-on-flat reciprocating, WC ball, 2 N, 25 m) were analyzed using friction curves and wear volume measurements, supported by 3D wear-track profiling and SEM investigations of wear tracks. Rapid T6 routes produced a refined, homogeneous distribution of spheroidized Si, with hardness higher than that of conventional T6. The rapid T6 routes using 5 min solutionization showed the lowest wear volume, whereas conventional T6 exhibited ~ 50% greater wear volume loss than the as-built state due to Si coarsening and matrix softening. Almost all rapid T6 conditions, except 10 minutes at 510°C, improved wear resistance relative to conventional T6, and the optimal rapid T6 route (540°C, 5 minutes) achieved approximately 40% better wear performance. In general, the wear damage was primarily caused by oxidation and abrasion. This study demonstrates that rapid T6 provides a time- and energy-efficient post-processing route that improves wear resistance against sliding wear conditions.

1. Introduction

Among aluminum alloys, silicon alloys are frequently preferred where wear resistance is required. With the weight-reduction potential of additive manufacturing technologies, Al-Si components enable the production of complex-shaped parts across the transportation, energy, aviation, and military sectors [1], [2], [3]. In additive manufacturing, the melt pool's laser efficiency and fluidity directly affect the total quality of the materials produced. The silicon content of the aluminum powder raw material has a positive effect on both parameters. In addition, thanks to the magnesium in the AlSi10Mg alloy, the parts produced are precipitation-hardened by heat treatment (with β -Mg₂Si precursor phases) [4], [5]. High cooling rates (between 10⁵ and 10⁸ K/s) in additive manufacturing lead to non-equilibrium solidification, and as the silicon network forms, supersaturated α -Al microstructures are produced [6]. Due to this microstructure, various hardening mechanisms (solid solution hardening, microstructural refinement, precipitation hardening, and aggregated second phases) can be applied to the additively manufactured (AM) AlSi10Mg alloy. Thanks to these secondary processes, harder and stronger products can be obtained from Al-Si casting alloys [7].

The precipitation mechanism in AlSi10Mg aluminium alloy is given below [8]:

Supersaturated solid solution (SSS) → Atomic clusters (Si and Mg atoms) → Guiner-Preston (GP) zones → β'' → β' → β .

New heat treatments, such as direct ageing, T6 heat treatment, hot isostatic pressing (HIP), or friction stir processing (FSP), are being investigated to improve the properties of additively manufactured

AlSi10Mg alloys [1], [9], [10], [11], [12]. One of the most common and beneficial heat treatments for aluminium alloys is T6. The T6 heat treatment is widely used, especially for fatigue strength. There are many different versions, but the most common parameters are solution heat treatment (SHT) at 540°C for 2 h, followed by water quenching (WQ) and then artificial ageing at 160°C for 12 h [13], [14]. T6 heat treatment is applied at high temperature and for a long time. In addition, long-term heat treatment, or HIP, will lead to higher energy consumption, lower productivity, and resource waste [15]. The United Nations Framework Convention on Climate Change focuses on reducing emissions in many industrial sectors. With this motivation, additive manufacturing of material process parameters and customized heat-treatment studies are becoming important [16]. While developing new-generation heat-treatment techniques that are environmentally sustainable and cost-effective, some studies have identified ways to mitigate the adverse effects of long-term, high-temperature heat treatments [17], [18].

In wear studies of as-built (AB) AlSi10Mg alloys produced by additive manufacturing, examinations were mainly conducted in the as-built condition, and the AB alloys were found to exhibit better wear resistance than their cast counterparts [19], [20], [21], [22], [23], [24], [25], [26]. The friction coefficient of AB AlSi10Mg material was found to be higher than that of the cast version [21]. Moreover, some studies reported that the onset of abrasive wear is triggered by hard Si phases and Al_2O_3 debris formed on the wear track, in addition to adhesive wear and tribooxidation phenomena typical of AlSi cast alloys [21], [23], [26].

Although AB L-PBF Al-Si alloys have good wear resistance, they are weak in tensile strength and ductility. Some heat treatments are performed to overcome this weakness [7]. In particular, two most common heat treatments are performed for AlSi10Mg alloys: (i) direct artificial ageing (T5), which precipitates Si nanoparticles and $\beta\text{-Mg}_2\text{Si}$ precursor phases and has a positive effect on ductility, strength and residual stresses [27]; (ii) solution treatment (T6), which, in addition to precipitation hardening with $\beta\text{-Mg}_2\text{Si}$ precursor phases, forms a composite-like structure with spherical Si particles embedded in the $\alpha\text{-Al}$ matrix and provides a good combination of mechanical properties [17]. Some studies have investigated the relationship between the mentioned heat treatments and wear resistance. Wei et al. found that both the friction and wear behaviour of T6-treated AlSi10Mg alloy were negative, and that the wear rate increased proportionally with solution temperature and Si particle size [26]. In addition, Prashanth et al. reported that the wear rate increased proportionally with the heat treatment temperature [25]. Inversely, Park et al. reported that fine Si particles bonded to the $\alpha\text{-Al}$ matrix more strongly than the fibrous eutectic-Si network after T6, resulting in higher wear resistance than the AB condition [28].

A new heat treatment for AM AlSi10Mg alloys is proposed. A shorter solution treatment is proposed to reduce the adverse effects of gas trapped in the pores of AM AlSi10Mg material. In contrast, rapid solution treatment of AM AlSi10Mg alloy has been found to offer some advantages[29], [30]: (i) improvement of Si particle properties (size, morphology and distribution), (ii) reduction of volumetric expansion of gas pores, and (iii) increase in supersaturation of Al matrix and consequent acceleration of precipitation hardening kinetics. A higher Orowan strengthening effect and improved mechanical properties were observed with rapid T6 heat treatment, resulting from a more homogeneous distribution

of smaller Si particles [17]. Thanks to this approach, production times were reduced, and mechanical properties were improved. This confirmed the applicability of this innovative T6 solution treatment on different Al-Si alloys [29].

Therefore, this work aims to investigate the influence of the innovative T6 heat treatment on the tribological behaviour of the AM AlSi10Mg alloy by comparing the following conditions: AB, conventional T6, and rapid T6 (four different routes, as described in the experimental section). These findings are considered of significant industrial importance because AlSi10Mg alloys are used in critical areas today.

2. Materials and Methods

The EOS M290 system was used to produce AlSi10Mg parts. This study used AlSi10Mg powder produced by EOS. The chemical analysis of the powder used is given in Table 1.

Table 1
Chemical composition of AlSi10Mg powder (wt.%).

Si	Fe	Cu	Mn	Mg	Ni	Zn	Pb	Ti	Al
9.8	0.14	0.02	< 0.01	0.36	0.01	< 0.01	< 0.01	0.01	Rest

Samples were produced using the EOS M290 metal system via direct metal laser sintering (DMLS). The production parameters are given in Table 2.

Table 2
DMLS parameters.

Laser power (W)	Scan speed (mm/s)	Hatch distance (mm)	Layer thickness (μ)	Energy density (J/mm ³)
270	1300	0.19	30	36.43

The energy density was calculated according to the formula [31],

$$E = \frac{P}{v \cdot h \cdot t}$$

1

In this formula, E indicates the energy density (J/mm³); P, laser power (W); v, scan speed (mm/s); h is the hatch distance (mm), and t shows the layer thickness (mm).

The samples were heat-treated in a Nabertherm LH216/12 furnace under an argon atmosphere. Heat treatment parameters are given in Table 3.

Table 3
Heat treatment conditions.

Sample Code	Heat treatment process
AB	No treatment
RT6-1	510°C/5 min + WQ + 160°C/12h
RT6-2	510°C/10 min + WQ + 160°C/12h
RT6-3	540°C/5 min + WQ + 160°C/12h
RT6-4	540°C/10 min + WQ + 160°C/12h
T6	540°C/2h + WQ + 160°C/12h

Metallographic examinations were performed using a Zeiss Axio Lab. A1 light metal microscope and a Hitachi SU3500 T2 scanning electron microscope (SEM). The Thermo Fisher Helios 5CX FIB-SEM device was used for detailed microstructure analysis. The EDAX Elite Super device conducted an Energy-Dispersive X-ray Spectroscopy (EDX) analysis. X-ray diffraction (XRD) measurements were performed with a step size of 0.0131° over 2θ from 20 to 90° . The operating conditions of the Malvern Panalytical Emp3 model X-ray diffractometer with Cu- α radiation ($\lambda = 1.541874 \text{ \AA}$) were 45 kV and 40 mA. Porosity analysis was performed on a Zeiss Smartzoom5 using the Zen Core 3.5 software. Porosity measurements were performed using a colouring technique.

Wear studies were performed using a Tribotech linear wear setup. Before all wear tests, the sample surfaces were ground and polished to a mirror-like finish. The experimental parameters were a WC ball with a 6 mm diameter, a 2 N load, a maximum sliding speed of 10 mm/s, a 25 m sliding distance, a 5 mm sliding amplitude, a $23 \pm 1^\circ \text{C}$ temperature, a $45 \pm 2\%$ relative humidity, and a 1 Hz data acquisition rate. Three tests were applied to each sample to ensure the reproducibility and validity of all wear tests, and the average measurements were presented.

3. Results and Discussion

Figure 1 shows the optical microstructure images of the samples. The As-Built microstructure consists of a Melt Pools (MPs) pattern with depth and shape differences (Fig. 1-f). The intergrown structure is formed by a 67° counter-clockwise rotation of the scanning strategy, which favors a more isotropic microstructure [32], [33]. RT6 heat treatments eliminate the MP structure and result in a composite-like microstructure consisting of a homogeneous distribution of Si particles embedded in the α -Al matrix (Fig. 1(a-e)). As reported in previous studies, the solution step of the T6 process results in the breakdown of the eutectic-Si network and spheroidization of the Si phase, resulting in a complete microstructural rearrangement [17], [18]. As reported in the literature, the size of the Si particles embedded in the α -Al matrix increases with increasing solution temperature and time [17]. For extremely short solution-temperature time, there is a more homogeneous distribution of smaller Si particles, which

contributes to higher adhesion between the Al matrix and Si particles and to a better balance between strength, ductility, and toughness [5], [7].

Figure 2 shows SEM micrographs of as-built samples, with the eutectic-Si network patterns clearly visible. These eutectic-Si network structures are a result of additive manufacturing [34].

Diffusional mechanisms induced by heat treatments cause the eutectic-Si network to break down and for Si precipitates to become apparent (Fig. 3). Increasing heat treatment time and temperature trigger the coarsening of Si particles embedded in the α -Al matrix. There is a transition from the order of many small, close-together Si particles to the order of a few large, far-from-each-other Si particles. It was determined that the distance between particles increased from 0.39 μm to 1.79 μm (RT6-1: 0.39, RT6-2: 0.61, RT6-3: 1.17, RT6-4: 1.32, T6: 1.79).

The Authors evaluated the tensile properties (yield strength (YS), ultimate tensile strength (UTS), and elongation to failure (ef)) of the AB and heat-treated L-PBF AlSi10Mg alloys, which are reported in Table 4. The RT6-4 sample exhibits the best mechanical properties. Previous studies have shown that all mechanical properties are improved compared to the conventional T6 heat treatment due to a more homogeneous distribution of fine Si particles in the structure after rapid T6 heat treatment [26], [35]. The modified Si particles prevent the initiation of fracture failure. Rapid T6 treatments have significantly improved ductility while achieving UTS and YS values similar to those of conventional T6. It is thought that the fine Si particle/ α -Al matrix interface promotes the merging of voids and achieves higher elongation values [25], [35].

Looking at the material porosity ratios given in Table 4, the lowest porosity was observed in the as-built sample. The highest porosity ratio was observed in the T6 heat-treated sample, which had the longest heat-treatment time and the highest temperature. In the rapid T6-treated samples, the porosity ratio generally increases with increasing treatment temperature and time. However, there is no consensus in the articles examining the relationship between heat treatment and porosity. Nevertheless, some claims put forward in the literature regarding the porosity-thermal load relationship are:

- It has been stated that when the laser beam contacts the powdered metal, the gases trapped within it cannot escape due to rapid cooling, and that their volume expands after heat treatment [36], [37].
- It has been noted that the protective atmospheres used during the process become trapped within the structure and expand under high temperature and pressure conditions [17], [38].

Table 4
Static mechanical properties as a function of heat treatment conditions.

Sample Code	YS (MPa)	UTS (MPa)	e_f (%)	Porosity (%)
AB	389 ± 8	444 ± 11	6.11 ± 0.5	0.59
RT6-1	268 ± 4	316 ± 3	12.03 ± 0.8	0.91
RT6-2	265 ± 5	308 ± 3	13.10 ± 0.6	1.01
RT6-3	279 ± 5	335 ± 5	14.6 ± 0.3	0.98
RT6-4	281 ± 6	339 ± 4	14.71 ± 0.3	1.02
T6	267 ± 6	300 ± 8	9.09 ± 0.4	1.25

Figure 4 presents the XRD spectra for the as-built and heat-treated AlSi10Mg samples. The heat-treatment parameters and substrate materials used in the detailed XRD analyses reported in the authors' previous studies are identical to those used in this study. In the previous article, Al and Si peaks were clearly detected; the Mg_2Si precipitates remain amorphous or too fine to detect, consistent with prior reports [39], [40], [41]. The dissolution of silicon in the aluminum matrix, achieved through heat treatment, resulted in increased Si peaks with increasing time and temperature [42]. Apart from the appearance of Si peaks, no new phases were observed, which indicated that both the rapid T6 route and the conventional T6 route simply produced a composite-like microstructure consisting of α -Al and globular Si [35].

According to the hardness results (Fig. 5), heat treatment significantly affected the alloy's strength. As expected, the LPBF AlSi10Mg sample, with its ultrafine cellular structure and continuous Si network, exhibited the highest hardness [35]. Based on previous studies, microhardness was reported at 120 to 130 HV in the as-built state [28]. Conventional T6 treatment resulted in a significant decrease in hardness, as shown in Fig. 5, consistent with dissolution of the fine eutectic Si network and coarsening of Si particles. A similar reduction in microhardness has been reported in the literature after a conventional T6; for instance, hardness dropped from 125 HV (as-built) to 85–100 HV [28], [43]. As shown in Fig. 5, rapid T6 treatments mitigated softening of the α -Al matrix and, as a result, all the rapid solution-treated samples (RT6-1 to RT6-4) retained higher hardness than conventional T6, the hardness slightly increased as the solution temperature increased. It is suggested that a rapid solution at 510–540°C was sufficient to re-spheroidize and redistribute Si, but avoided the extensive Si growth and over-aging observed in the conventional T6 treatment. Moreover, a shorter solution retained a higher supersaturation of Si in the matrix, enhancing subsequent precipitation hardening during aging [17].

The sliding-friction responses of the various samples are compared in Fig. 6 (COF vs. sliding distance curves) and Fig. 7 (mean and maximum COF values). All samples exhibited an initial run-in period followed by a steady-state friction regime, but steady-state COF levels differed across conditions. The as-built sample showed a relatively low COF during sliding action and the COF curve was more stable

and continuous compared to the heat-treated conditions, which exhibited wider oscillations or more irregular fluctuations in the COF evolution during the sliding tests. Thasleem et al. [21] observed that the heat-treated microstructure of SLM-processed AlSi10Mg alloy led to a higher COF (in dry sliding against steel) than the as-built SLM-processed AlSi10Mg alloy. In contrast to rapid-heat-treated samples, the conventional T6-treated sample generally exhibited lower, more stable COF curves after run-in. The conventional T6 sample showed the lowest mean and maximum COF for all heat treatment conditions (Fig. 7). Its steady-state COF is slightly above that of the as-built sample, suggesting that the removal of the continuous Si network and the presence of a softer α -Al matrix allowed a more conformal contact or more effective lubricious oxide layer formation than rapid heat-treated samples.

The cumulative wear volume of AlSi10Mg samples under various conditions is compared in Fig. 8. It was found that the RT6-3 condition resulted in the lowest wear volume loss (0.16518 mm^3), which indicates superior wear resistance, while the RT6-2 condition showed the highest wear volume loss (0.4593 mm^3), nearly three times greater than the RT6-3 condition. The AB sample showed an intermediate wear volume loss (0.23996 mm^3), whereas the conventional T6 sample showed a higher wear volume loss (0.35454 mm^3). Other rapid treatments fall within the range of RT6-1: 0.22328 mm^3 and RT6-4: 0.31215 mm^3 . These results demonstrated that appropriate heat treatment can markedly improve wear resistance. Specifically, the RT6-3 treatment improved wear performance by $\sim 30\%$ relative to AB. In contrast, a detrimental treatment like RT6-2 can nearly double wear volume loss.

The underlying hardness and microstructural changes caused by heat treatment can be directly linked to these wear trends. The hardness of a material is one of the primary factors that affect wear, as the hardness increases, wear loss tends to decrease. The as-built SLM material has a very fine microstructure (an ultrafine cellular α -Al matrix with a continuous Si-rich network at cell boundaries), resulting in a hardness of $\sim 120 \text{ HV}$. This explains why the AB sample already showed a reasonably low wear volume loss. By contrast, conventional T6 heat treatment (solution + full aging) coarsened the microstructure significantly as the fine Si network was dissolved and re-precipitated as coarser ($\sim 4 \mu\text{m}$) Si particles, and Mg_2Si precipitates formed in the matrix [26]. This coarsening and phase redistribution lowered hardness. The rapid T6 heat treatment was designed to refine this outcome by adjusting the thermal process, and the wear test results confirmed their impact. In RT6-3 as the best-performing condition, the treatment likely achieved an optimal balance between sufficient precipitation of fine strengthening phases without excessive coarsening of Si. Aging at an optimal temperature/time can densely precipitate nanoscale Mg_2Si (β'') in the α -Al cells while only spheroidizing or partially reducing the eutectic Si network. This leads to a hardened microstructure (approaching or even exceeding the as-built hardness) and thus minimal wear. In contrast, the RT6-2 appeared to produce a relatively softened microstructure. The other two samples (RT6-1 and RT6-4) showed intermediate wear behavior, suggesting they achieved only partial improvements.

As shown in Fig. 9, 3D profiles of the wear tracks provide both visual and quantitative confirmation of differences in wear volume loss between the samples. There were significant differences in the track

depth and width of the worn surfaces, which corresponded to the volume of material removed. Among all conditions, the RT6-3 sample has the shallowest and narrowest wear track, consistent with its minimal wear volume. The RT6-2 sample showed a much deeper, wider scar due to the large volume of material lost. A deeper or wider wear scar indicated a larger cross-sectional area of material being removed, hence a higher wear volume loss for a given sliding distance. In contrast, the wear tracks of the AB and RT6-1 samples were relatively shallow (only a mild penetration into the surface), while those of the T6 and RT6-4 samples were deeper and broader, though still less severe than those of the RT6-2 sample.

The RT6-2 sample exhibited extensive grooving and a wide crater, both of which indicate severe damage. As the material yields under load, a softer or heavily worn sample will produce a wider contact patch (wider track), whereas a harder sample will confine the contact and wear to a smaller area. As a result of these 3D profiles, the quantitative data were validated (the smallest cross-sectional area for the RT6-3 sample corresponds to the lowest wear volume loss, and the largest for the RT6-2 sample corresponds to the highest wear volume loss) and visually illustrated the advantages of limiting wear scar dimensions through a hard, strengthened microstructure. Additionally, the shallower depths of the best samples suggested that more sliding was carried out on tribofilms (discussed below), whereas the deep track of the worst sample indicated that metal-to-metal contact occurred and severe material losses occurred.

SEM (Fig. 10) and EDS (not shown here) examination of the wear tracks revealed that all conditions were subject to oxidative wear to some extent, as confirmed by the formation of oxygen-rich tribolayers on the worn surfaces. For the as-built sample (Fig. 10a), EDS analyses measured about 60.7 at.% O together with 35.9 at.% Al and 3.4 at.% Si that indicated the formation of an oxide layer composed mainly of Al oxide. This oxygen-rich surface film accounted for the relatively mild wear morphology observed in the as-built condition, which consisted of a smooth wear track with thin oxide patches and fine parallel grooves resulting from limited micro-abrasion. A similar response was observed for RT6-1 (Fig. 10b), with the worn surface showing shallow ploughing marks and relatively continuous oxide coverage. This indicated that oxidative wear remained dominant and abrasive damage was limited.

In contrast, RT6-2 (Fig. 10c) exhibited the most severe surface degradation. SEM observations revealed deep grooves, torn fragments, smeared material, and clear signs of delamination, all of which indicate severe adhesive wear combined with abrasive cutting. Although oxygen was still detected by the EDS on the wear scar, the oxide film in this condition was evidently unstable and unable to protect the surface under load. The softened over-aged microstructure promoted breakdown of the tribolayer, direct metal-to-metal contact, and subsequent material transfer and pull-out, leading to the highest wear volume loss [44]. The conventional T6 sample showed intermediate behavior. There were oxide fragments in the wear track, along with scratches, scratch marks, and local craters, suggesting that oxidative wear was present but accompanied by adhesive and abrasive wear indicators. It is thought that this mixed mechanism is due to coarsened Si particles and reduced continuity of the protective surface film.

Among all conditions, RT6-3 (Fig. 10d) displayed the most favorable wear morphology. Its wear track was largely smooth and flat, with only very shallow grooves and extensive coverage by a thin, adherent oxide layer. Based on these characteristics, mild oxidative wear was the dominant mechanism, and adhesion and abrasion were strongly suppressed. The improved performance of RT6-3 can be attributed to its optimized microstructure, which is believed to support the formation and retention of a stable tribo-oxide film while limiting plastic deformation during sliding. Although RT6-4 (Fig. 10e) showed predominantly oxidative wear, its wear track contained some exposed metallic regions and slightly deeper grooves than RT6-3, which may indicate partial cracking or local breakdown of the oxide film.

Overall, the SEM and EDS results demonstrated that wear performance was influenced by the balance between protective oxidative wear and destructive adhesive-abrasive wear. Oxygen-rich tribolayers formed in all samples, but high wear resistance was achieved only when this oxide film remained continuous and adherent during sliding. Conditions such as the as-built state, RT6-1, and especially RT6-3 maintained more stable oxide coverage and therefore showed smoother tracks and lower material loss. By contrast, RT6-2 and T6 (Fig. 10f) samples were affected by repeated oxide breakdown, severe grooving, delamination, and metal transfer, which caused substantially higher wear volume loss. These results indicated that the most effective tribological response was observed in RT6-3, where the combination of high hardness, a refined microstructure, and a stable oxide tribofilm promoted mild oxidative wear and minimized destructive surface damage.

4. Conclusions

The AB AlSi10Mg condition exhibited good wear resistance in dry sliding due to its ultrafine cellular microstructure with a continuous Si-rich network and the highest hardness, resulting in the low wear volume and the shallow wear track. In contrast, conventional T6 produced poor tribological performance, with a wear volume approximately 1.5x higher than AB and significantly deeper/wider tracks. This deterioration can be directly attributed to solution-treatment-induced microstructural coarsening. The continuous Si network transformed into coarse, isolated Si particles ($\sim 1-3\ \mu\text{m}$), accompanied by significant softening of the α -Al matrix. Moreover, prolonged high-temperature exposure may also promote porosity-related damage during sliding.

Almost all rapid T6 conditions, except for 10 minutes at 510°C, substantially outperformed conventional T6 by limiting Si coarsening and preserving a composite-like architecture of fine, homogeneously distributed Si particles within a strengthened matrix. As a result, rapid T6 samples retained a high level of hardness, with the 540°C/5 min condition providing the best wear resistance among all heat-treated states. Overall, optimized rapid T6 conditions provide an energy- and time-efficient post-processing strategy that offers better wear resistance than AB and conventional T6 heat-treatment routes, as well as improved mechanical performance, particularly important for wear-critical and load-bearing AM AlSi10Mg components.

Declarations

Declaration of Competing Interest

The authors declare that they have no known competing financial interests or personal relationships that could have influenced the work reported in this paper.

Author Contribution

Mustafa Safa Yılmaz: Formal analysis, Methodology, Validation, Writing-original draft, Faiz Muhaffel: Formal analysis, Investigation, Writing-original draft, İbrahim Tütük: Investigation, Data curation, Methodology, Gökhan Özer: Resources, Writing – review & editing, Methodology, Conceptualization, Funding acquisition.

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Figures

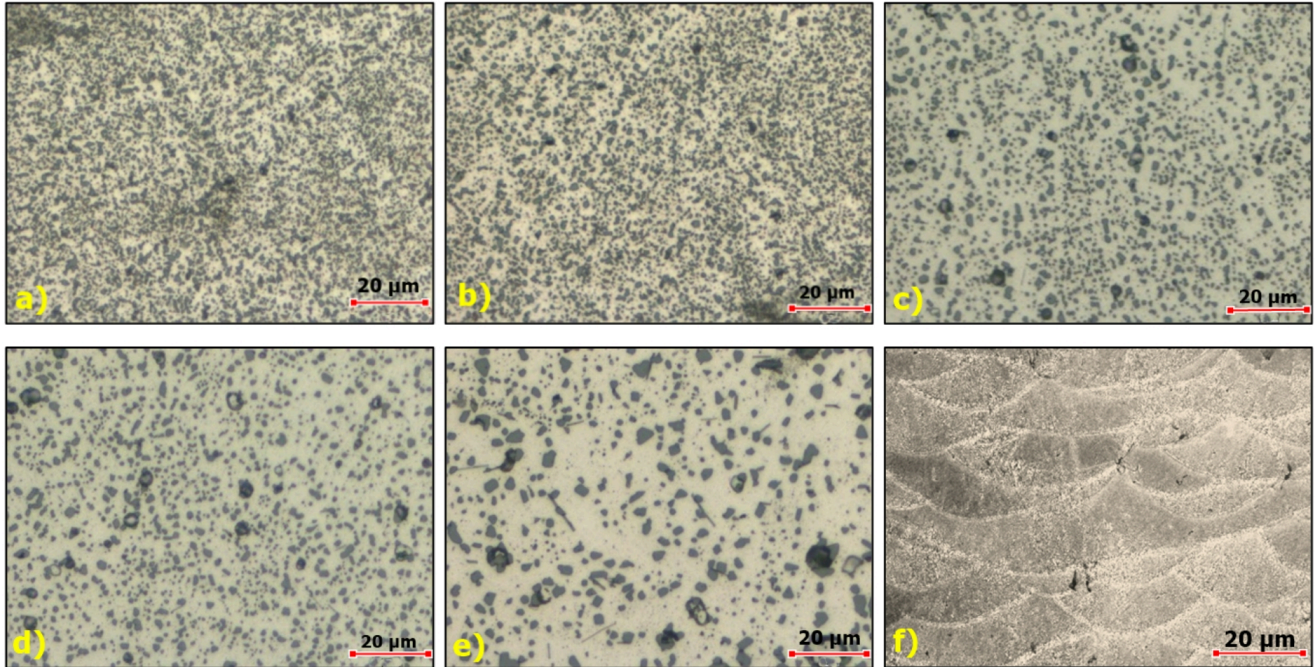


Figure 1

Optical microscope images of the samples, (a) RT6-1, (b) RT6-2, (c) RT6-3, (d) RT6-4, (e) T6, (f) as-built.

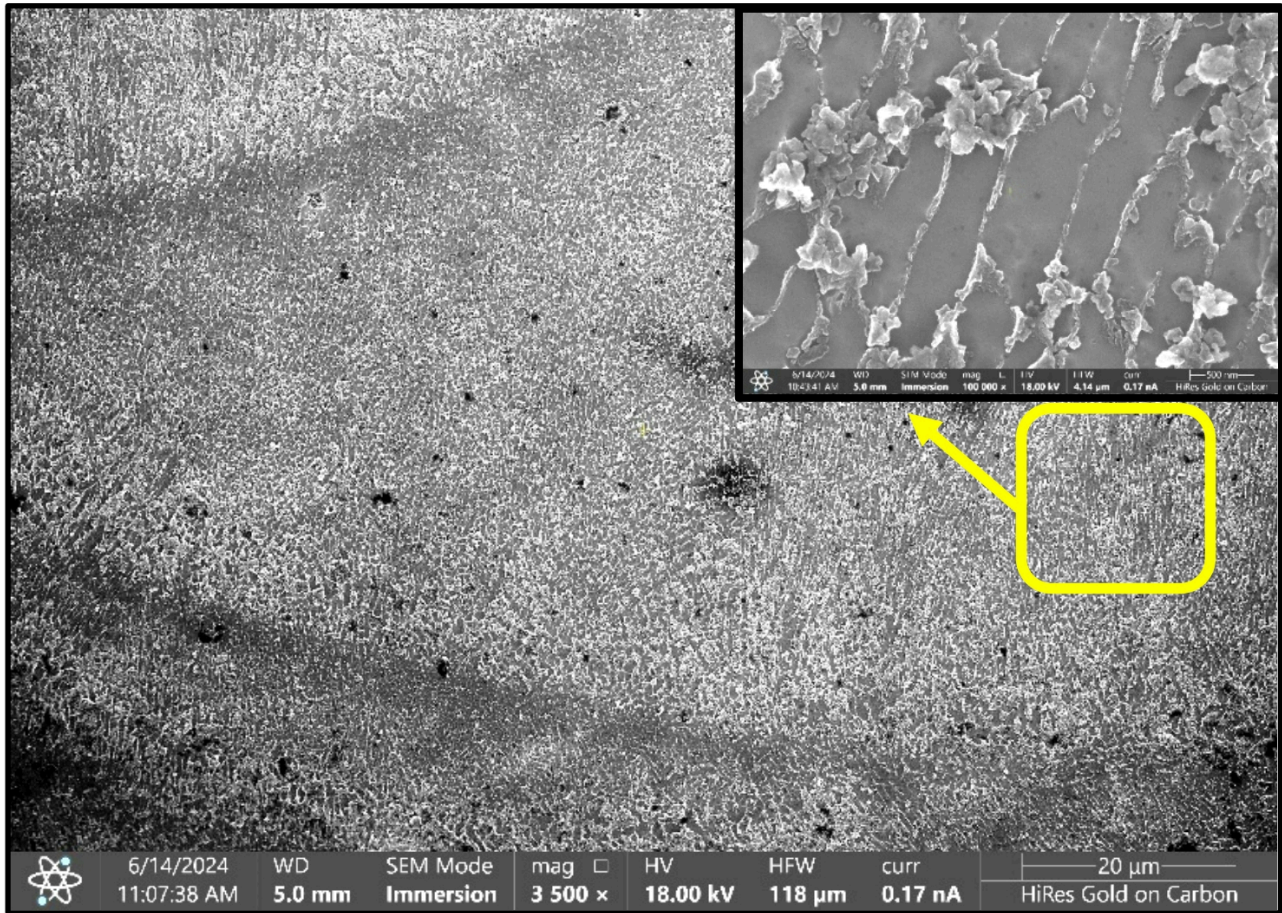


Figure 2

SEM images of the as-built sample.

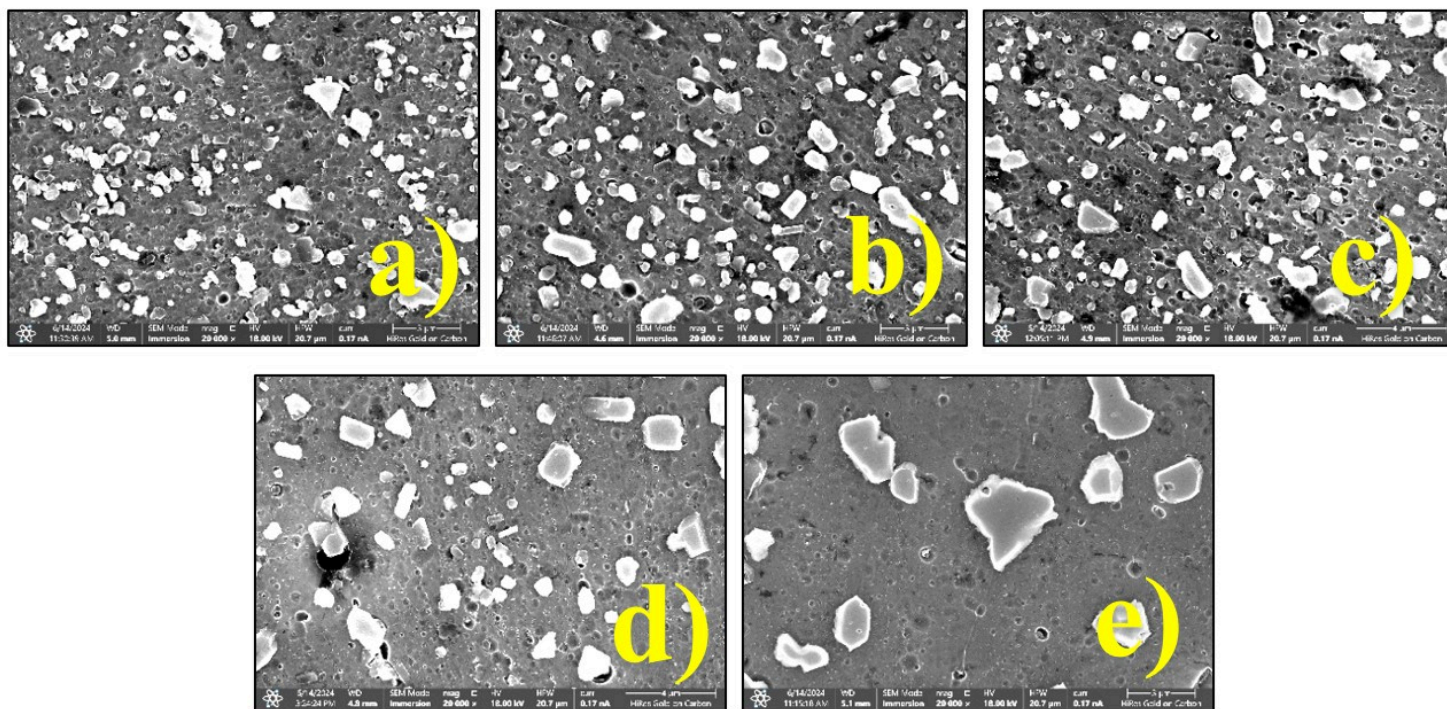


Figure 3

SEM images of the samples, (a) RT6-1, (b) RT6-2, (c) RT6-3, (d) RT6-4 and (e) T6.

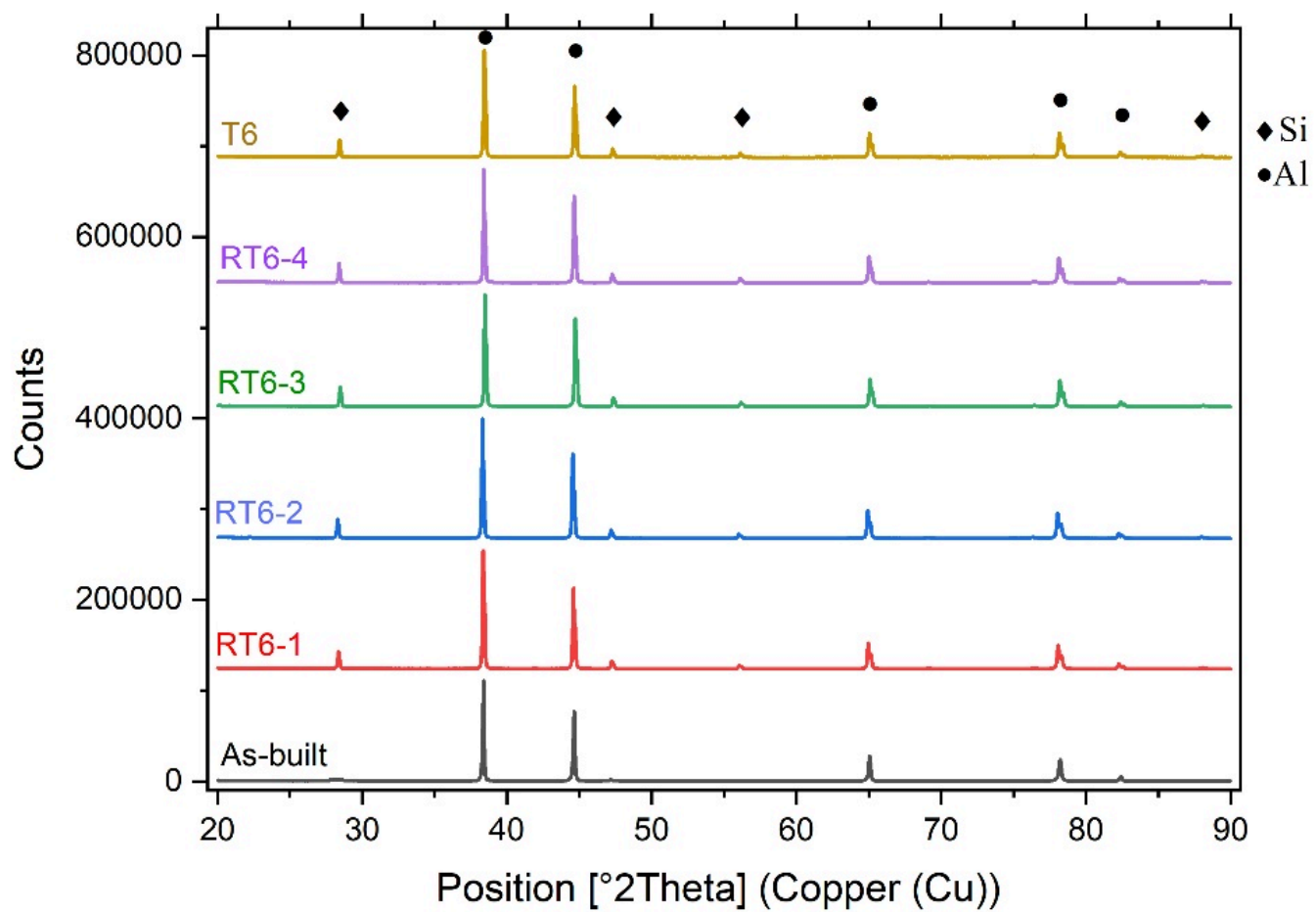


Figure 4

XRD spectra of the samples.

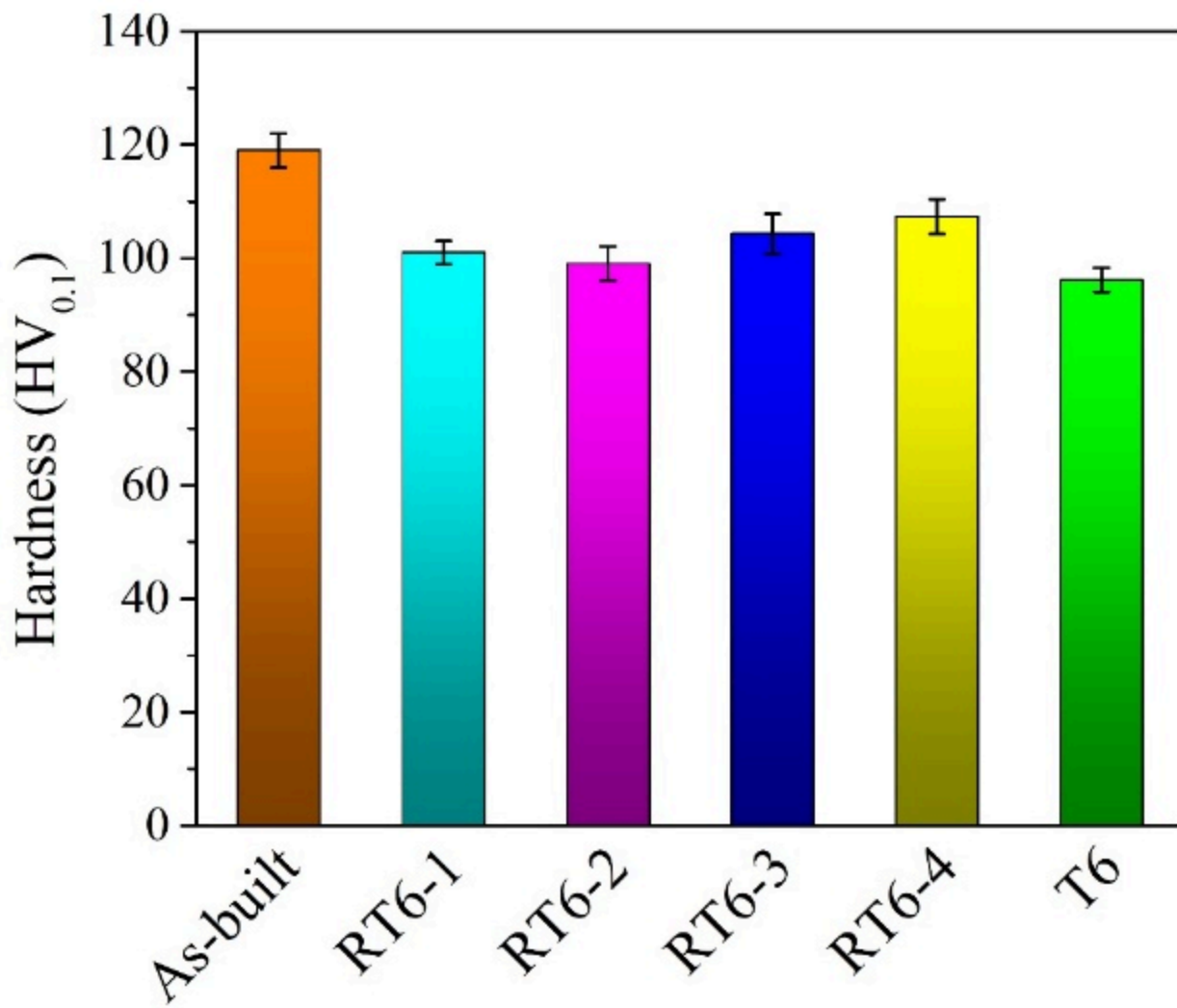


Figure 5

Hardness of the samples.

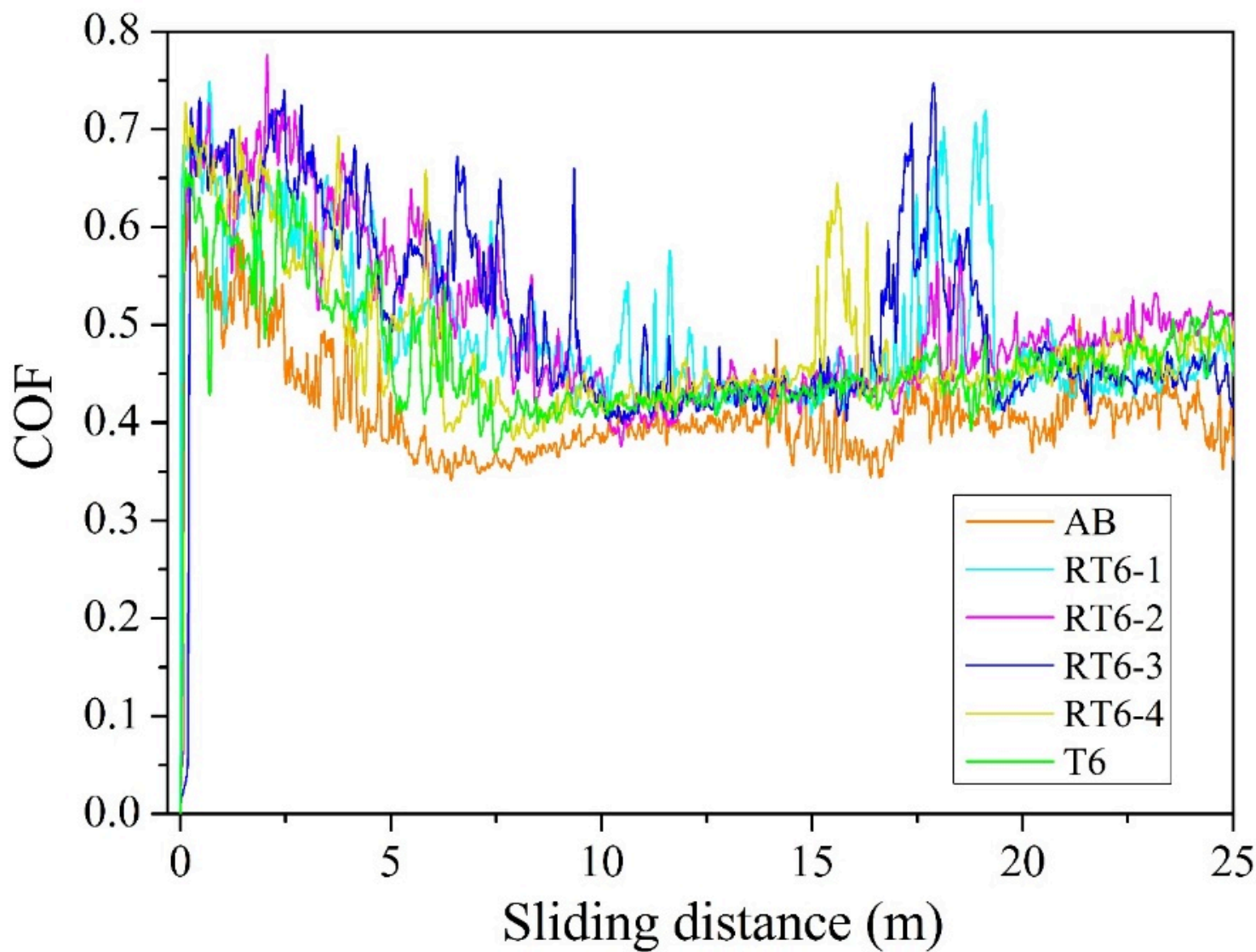


Figure 6

COF curves of the samples.

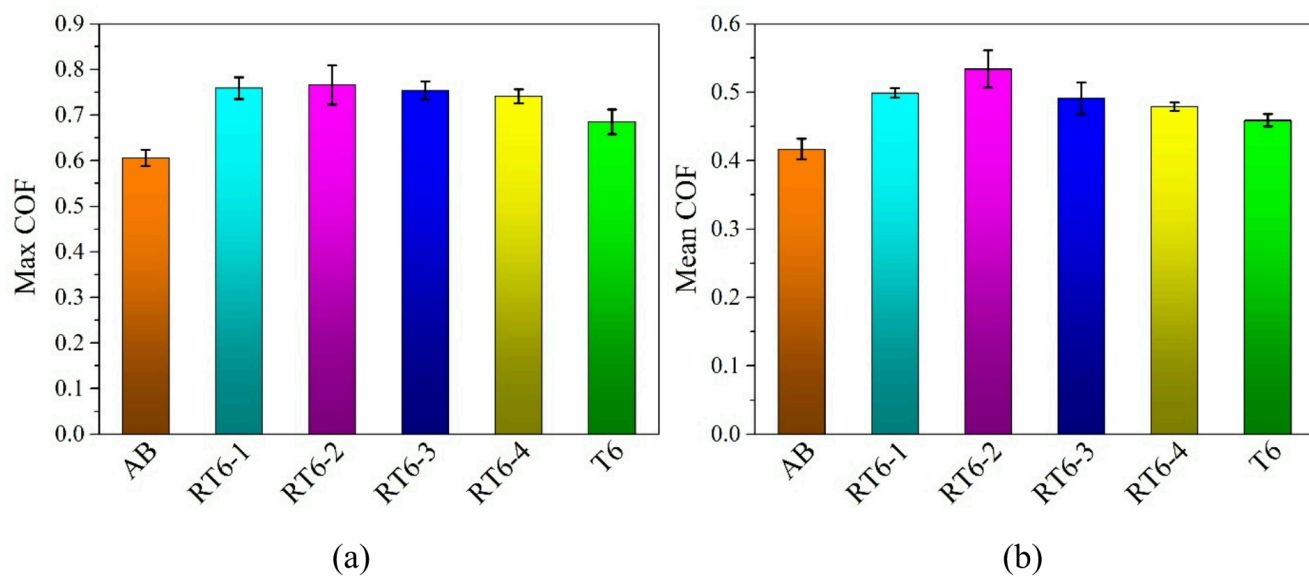


Figure 7

Max and mean COF values of the samples.

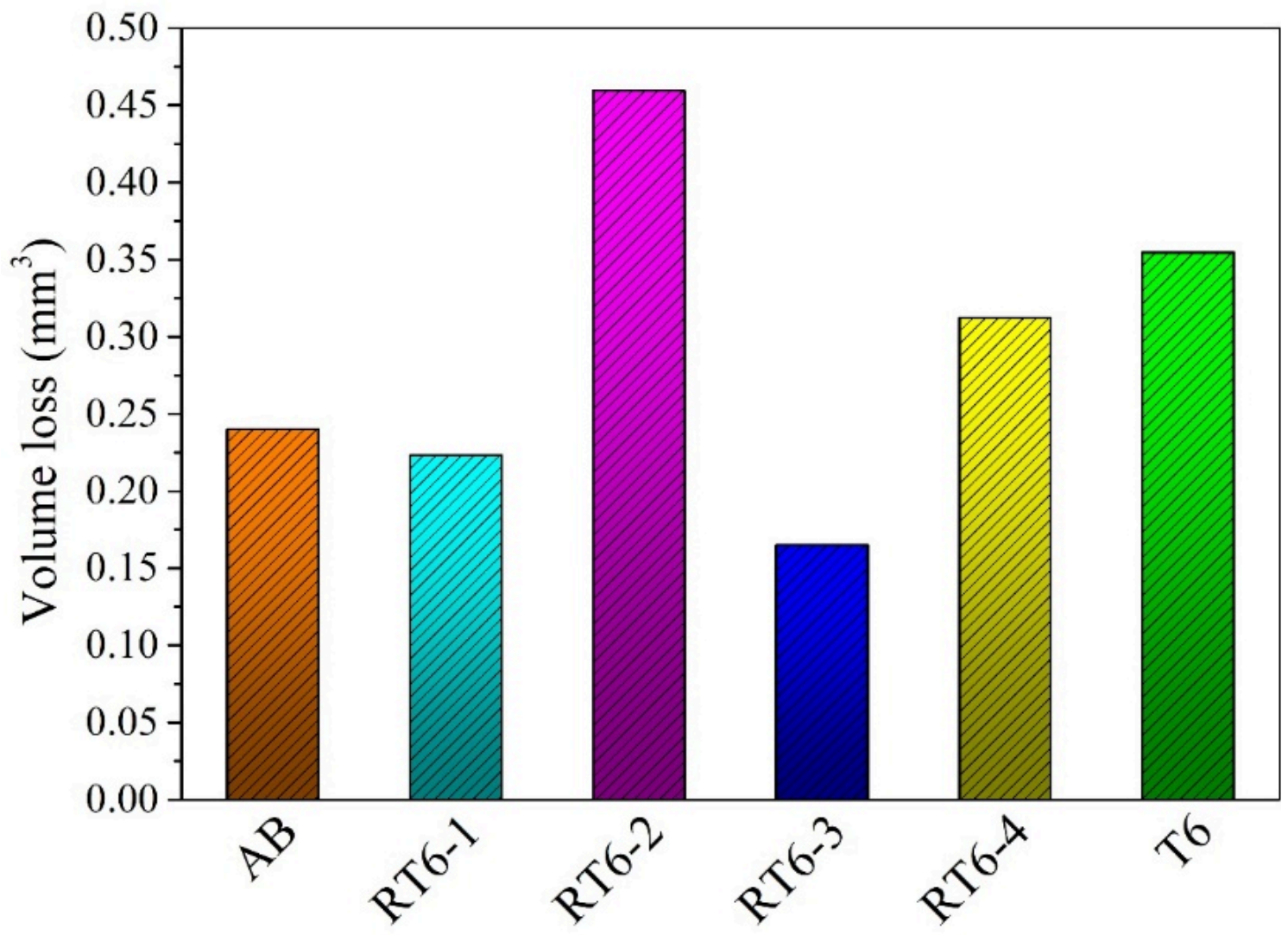


Figure 8

Wear volume of the samples.

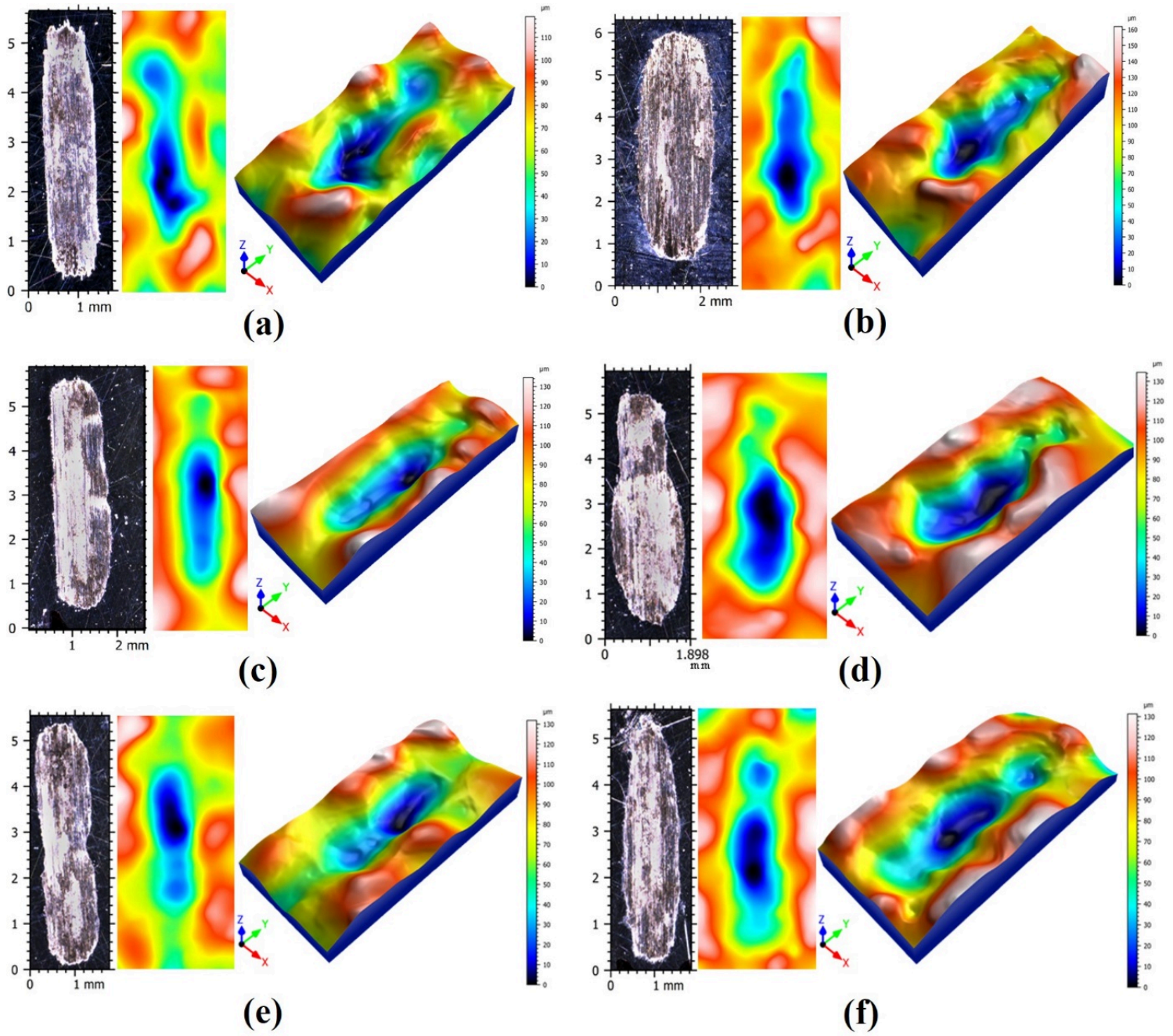


Figure 9

Wear track 3D profiles of the samples. (a) as-built, (b) T6, (c) RT6-1, (d) RT6-2, (e) RT6-3, (f) RT6-4.