Enhanced Cavitation Erosion Resistance of Friction-Stir Processed High Entropy Alloy

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Abstract: In the present work, friction stir processing of Al_{0.1}CoCrFeNi high entropy alloy (HEA) was performed at controlled cooling conditions (ambient and liquid submerged). Microstructural and mechanical characterization of the processed and unprocessed HEAs was evaluated using electron-back scattering diffraction (EBSD) and micro and nanoindentation. HEA under submerged cooling condition showed elongated grains (10 µm) with fine equiaxed grains (2 µm) along the boundary compared to the coarser grain size (∼2 mm) of as-cast-HEA. Hardness showed remarkable improvement with four times (submerged cooling condition) and three times (ambient cooling condition) improvement than as-cast HEA (∼150 HV). Enhanced hardness is attributed to significant grain refinement for processed HEAs. Cavitation erosion was behavior of all the samples was evaluated using ultra sonication method. All the HEAs showed better cavitation erosion resistance than the stainless steel. The sample processed under submerged liquid condition showed around 20 and 2 times highest erosion resistance than SS316L and as-cast HEA, respectively. The enhanced erosion resistance of the processed HEA is related to their increased hardness, resistance to plasticity and better yield strength compared to the as-cast-HEA. Surface examination of the tested samples showed nucleation and growth of pits, and plastic deformation of material followed by fatigue-controlled disintegration are the primary material removal mechanism.

Keywords: Erosion, Microstructure, Mechanical properties, Surface Engineering

1. Introduction

Hydraulic machinery such as ship-propellers, pump-impellers, valves and hydro-turbines are severely affected by slurry and cavitation erosion leading to substantial financial losses [1-3]. Cavitation erosion is a serious concern formed due to the implosion of vapor bubbles resulting from rapid pressure fluctuations in the liquid. During implosion, high intensity shockwaves and micro-jets are produced, which impacts on the solid surface at a high frequency. This repeated impact of micro-jets and shockwaves leads to the formation of higher density pits, plastic deformation and eventually material failure [1]. The rate of material degradation is more aggressive
if cavitation erosion occurs in a corrosive environment. State-of-the-art materials used in these applications showed limited resistance to cavitation erosion. Therefore, advanced materials and surface modification techniques are essential to counter the degradation phenomenon.

Recently developed high entropy alloy (HEAs) received considerable attention as structured materials due to its promising properties and unique microstructural features [4, 5]. Unlike conventional alloys, HEAs are composed of at least five major elements in equimolar or nearly equimolar fractions. HEAs tend to form a single-phase solid solution with simple crystal structures such as face-centered cubic (FCC), body-centered cubic (BCC) and/or hexagonal closed pack (HCP) [4-8]. The formation of single-phase solid solution is due to large mixing entropy thereby minimizing the overall free energy and enhance the stability of the system [9]. Further, the thermodynamic parameters such as atomic size difference and mixing enthalpy also contribute to the formation of single-phase crystal structure. The highly distorted lattice and slower diffusion kinetics are likely expected due to the difference in atomic sizes and high configurational entropy [5]. Recent studies have shown that Al0.1CoCrFeNi and CoCrFeMnNi HEAs exhibit single-phase FCC structure with excellent mechanical properties. These HEAs were reported to have large work hardening capability, high ductility, excellent wear and oxidation resistance and good corrosion resistance [7, 10, 11].

So far, HEAs show better wear and corrosion resistance compared to conventional materials [12-14]. However, erosion-corrosion behavior of HEAs was likely unexplored to some extent. Recently we reported the Al0.1CoCrFeNi HEA shows excellent slurry erosion resistance to mild steel at a normal angle and similar erosion rate at an acute angle [15]. Also, the cavitation erosion-corrosion behavior of Al0.1CoCrFeNi HEA compared to stainless steel 316L [16]. Despite having lower hardness compared to SS316L, the cavitation erosion resistance was higher for the former. Better cavitation erosion performance of HEA is attributed to higher work hardening and excellent corrosion resistance. Higher erosion resistance is attributed to higher work hardening due to low stacking fault energy. Zhang et al [17] investigated the cavitation erosion behavior of laser surface alloyed AlCoCrFeNi HEA coating compared to stainless steel. HEA shows better cavitation erosion resistance and higher pitting resistance compared to stainless steel. AlCoCrFeNiTi2 was also reported with excellent cavitation erosion resistance due to the presence of B2 structure [17]. Although, the performance of HEAs have shown excellent improvement in cavitation erosion, the microstructural refinement further influences the tribological behaviour of the former. In particular, the effect of grain size as well as deformation induced phase transformation influencing cavitation erosion behaviour of HEA is still lacking.
Surface modification is the eminent method to tailor the microstructure. Various surface modification technique was introduced to improve the mechanical and tribological properties [18-20]. Surface modification through high strain deformation is eminent and cost-effective [21-23]. Friction stir processing (FSP) is thermo-mechanical processing used to tailor the microstructure by high strain deformation [2, 24, 25]. Although limited, few studies were reported related to friction stir processed HEAs. Kumar et al [26] investigated friction stir processed Al_{0.1}CoCrFeNi HEA under the ambient condition with 600 rpm rotational speed. Processed HEA shows significant improvement in strength and ductility with higher yield strength. Koraswamy et al [27] reported four times increment in yield strength for processed Al_{0.1}CoCrFeNi HEA by rotating at 1000 rpm with a 1.5 mm pin tool. However, the performance of FSPed HEA has not been evaluated under the cavitation erosion environment. Further, the correlation of cavitation erosion behavior with different mechanical properties has not been explored. In present work, we reported the cavitation erosion performance of friction stir processed Al_{0.1}CoCrFeNi high entropy alloy processed under different cooling conditions in comparison with as-cast HEA and stainless steel. The study showed bimodal (dual phase) HEA processed under submerged cooling condition showed superior cavitation erosion resistance.

2. Experimentation details

2.1. Materials and processing

The nominal composition of Al_{0.1}CrCoFeNi high entropy alloy was used in present work which has been reported earlier [15]. Friction stir processing (FSP) was conducted using a pin less tungsten carbide tool with a 16 mm shoulder diameter on a universal milling machine. The specially designed fixture was used for performing FSP. The parameters used for FSP were adopted from the previous study [28]. The FSP was performed at a constant rotational speed (388 RPM) and transverse speed (20 mm/min) under two different test conditions i.e. ambient cooling (referred as 388A) and submerged liquid cooling (referred as 388C). External cooling was provided for the sample performed under submerged conditions by maintaining a constant temperature of 0°C connected to chiller. The coolant used is the mixture of ethanol and distilled with equi proportions.

2.2. Microstructural and Mechanical characterization

The surface and cross-section of processed Al_{0.1}CoCrFeNi HEA samples were sectioned using a low precision diamond cutter. The sectioned samples were grounded down to 3000 grit followed by diamond polishing. The
polished samples were cleaned with acetone and dried, prior to characterization studies. The microstructural studies were conducted using electron backscatter diffraction (EBSD) analysis for the HEA samples. EBSD analysis was performed using a scan step of 0.1µm for processed HEA (FEI Quanta 3D FEG). Hardness along the cross-section for processed HEA was evaluated using micro-hardness testing (Wilson MV 402D) at a 50 gf load. Elastic modulus was also evaluated using the Oliver-Phar method from load-displacement curves [29]. Strain hardening exponent (n) for all the samples were obtained from nano-indentation testing (Hystrix TI 950).

2.2. Cavitation erosion testing

The cavitation erosion test was performed using ultrasonic vibratory apparatus according to ASTM G32 standard. The samples were kept stationary below the vibrating tip (indirect method) at a distance of 500 µm. The frequency of the vibrating tip was 20 ± 0.5 kHz with a peak-to-peak amplitude of 50 µm. Test samples were submerged in a 1-litre beaker containing test media (distilled water). To ensure the ambient temperature (24 ± 2°C) throughout the testing, the cooling coil connected to chiller was provided to the test media. Prior to cavitation erosion testing, all the tested samples (10 x 10 mm) were grounded down to 2000 grit using abrasive paper followed by cleaning using acetone. The samples were tested for 20 hours and subsequent mass loss (0.01 mg) was monitored for every one-hour cycle. The cumulative volume loss (CVL), cumulative erosion rate (CER) mean depth erosion rate (MDER) and incubation period were analysed to determine the cavitation erosion performance of the tested samples. The cavitation erosion resistance (Re) was evaluated by the reciprocal of MDER to determine the correlation with different mechanical properties. The morphology after cavitation erosion testing was analysed using scanning electron microscope (SEM) to evaluate the damage mechanism.

3. Results and Discussion

3.1. Microstructural characterization

Electron backscatter diffraction (EBSD) maps for as-cast and friction stir processed Al0.1CrCoFeNi HEAs are shown in Fig. 1. The as-cast HEA showed a coarser grain size around 2000 µm with a single-phase FCC structure without any secondary phase formation. Similar results were also reported for the Al0.1CoCrFeNi HEA with coarser grain size [30]. After FSP, the surface morphology of the HEA showed significant grain refinement, processed under ambient and submerged cooling conditions (Fig. 1 (b to f)). 388A HEA showed equiaxed grains with an average grain size of around 1.6 µm (Fig.1 (b and e)). On the other hand, the bimodal structure composed of fine equiaxed grains (≤2 µm) embedded in the coarser grains with a size around 10 µm was
observed for the 388C HEA (Fig. 1 (c and f)). The average grain size of the 388C HEA was around 7.2 µm. Our group also reported the development of bimodal structure (austenite and martensite) after friction stir processing for SS 316L [31]. After FSP, the variation in the texture was also observed. The as-cast HEA is predominantly composed of FCC phase with {100} being parallel to the surface. After processing a change in the texture was observed with mixed texture obtained for both the samples with presence of FCC and BCC grains. The 388A sample showed random texture while slightly coarser grains present in 388C sample predominantly retained the parent texture. Dynamic recrystallization (DRX) comprised of dislocation rearrangement and migrating boundaries is the predominant mechanism for microstructural refinement after friction stir processing [32]. DRX is strongly influenced by a material with low stacking fault energy (SFE). In contrast, HEAs also reported to have lower SFE due to the multi principal elements [33-35]. Zaddach et al [34] manifested the reduction in SFE due to increase in the number of elements. However, the SFE for the present HEA composition was observed to be less than 30 mJ/m² [36]. Eleti et al [37] reported very fine equiaxed grains along the boundaries for CoCrFeMnNi HEA during severe plastic deformation is attributed to DRX due to low SFE. The lower SFE is due to higher partial dislocations which further increase the dislocation density leading to DRX and eventually the grain refinement.

Further EBSD phase map reveals the formation of secondary phases in the HEA after processing (Fig. 2). In the case of 388A, a small fraction of BCC was observed (≈6%) while a higher fraction of BCC phase (≈21%) was observed for 388C. The BCC phase was observed at the grain boundaries of 388C. The difference in the phase fraction for 388C and 388A might be attributed to the difference in the cooling conditions during FSP. High strain-rate deformation during FSP results in enhancing the free energy of the system through higher defect density. Processing under ambient conditions (388A) allows the system to relax the stored energy through complete recrystallization. However, rapid cooling during FSP (388C) restricts the full relaxation of the stored free energy due to absence of temperature field. Instead, the system likely favors FCC to BCC transformation (strain induced transformation) to get to a more equilibrium state. Further, Tsai et al [38] observed slower kinetics of grain growth for CoCrFeMnNi HEA which is associated with higher activation energy and sluggish diffusion. Furthermore, our group study reported that the Al₀.₁CoCrFeNi HEA also showed higher activation energy which enhances the sluggish diffusion due to large fluctuation of lattice potential energy (LPE) sites [39]. The atoms get trapped at low LPE sites which hinders the diffusion of atoms leading to sluggish diffusion and thereby the new phase formation on HEAs after FSP.
Fig. 1. Electron backscattered images of (a and d) unprocessed, (b and e) 388 ambient cooling (388A) and (c and f) 388 liquid cooling (388C) of Al_{0.1}CoCrFeNi high entropy alloy.
3.2. Mechanical properties

The variation of hardness along the cross-section for processed Al\textsubscript{0.1}CoCrFeNi HEAs are shown in Fig. 3(a). The as-cast HEA showed an average hardness of around 150 HV which is comparatively lower than the stainless steel (227 HV). After processing, significant improvement in the hardness was observed for both the processed samples. The hardness showed a maximum at the top surface and gradually decreases throughout the depth for both the processed HEAs. The highest hardness (\approx 625 HV) was observed for 388C HEA with a nearly four-fold increment compared to as-cast HEA. 388A HEA showed around 3 times increment (\approx 429 HV) compared to unprocessed HEA. The higher hardness for the processed HEAs can be explained on the basis of significant grain refinement (grain boundary strengthening) and the presence of BCC phase embedded at the grain boundaries (388C). According to Hall-Petch relation, the lower grain size (higher grain boundaries) act as a barrier for dislocation motions which enhances the hardness [40]. Along with grain refinement, the presence of BCC precipitates at the grain boundaries also resulted in enhanced hardness. Similar studies reported for Al\textsubscript{0.25}CoCrFeNi HEA, which exhibits higher hardness after deformation [41]. The dislocations encountering the hard phase at the grain boundaries results in effective dislocation pinning and their intersections, which contributes to higher hardness for the processed HEAs. The higher fraction of hard precipitates (BCC) was observed for the 388C which explains the higher hardness compared to 388A HEA. The elastic modulus (\(E\)) for as-received and processed HEAs obtained from load-displacement curves using nano-indentation are shown in
Fig. 3(b). The $E$ value decreased for both the processed samples compared with as-cast HEA. The lower the elastic modulus for processed HEA is attributed to change in texture after friction stir processing. The yield strength for all the HEA samples calculated based on the equation given below

$$\sigma_y = \left( \frac{H}{n} \right) (0.1)^n$$

where, $H$ is the hardness, and $n$ is the strain hardening exponent [42]. The yield strength estimated for as-cast and processed HEAs are shown in Table 1. The average yield strength obtained for as received $\text{Al}_{0.1}\text{CoCrFeNi}$ HEA by the tensile test was around 160 MPa [15]. In contrast, both the values for as-cast HEA exhibited almost similar value. The 388A HEA showed yield strength around 367 MPa which around 2.3 times than as-cast HEA. The higher yield strength was observed for 388C (507 MPa) which is 3 times better than as-cast HEA. Further, the strain hardening exponent ($n$) was calculated using a model proposed by Giannakopoulos and Suresh [43]. Strain hardening exponent obtained for all the HEA samples are shown in Table 1. It is observed that the $n$ value slightly decreased for both processed HEA. Generally, grain boundary strengthening impedes dislocations motions results in decrease of strain hardening [44]. In case of 388C, the $n$ value was slightly higher compared to 388A which might be attributed to higher fractions of BCC phase.

Fig. 3. (a) Variation in the micro-hardness and (b) elastic modulus of as-cast and processed $\text{Al}_{0.1}\text{CoCrFeNi}$ high entropy alloy.
Table 1. The mechanical properties of as-cast and friction stir processed Al$_{0.1}$CoCrFeNi high entropy alloys in comparison with stainless steel 316L

<table>
<thead>
<tr>
<th>Samples</th>
<th>Hardness/[HV]</th>
<th>Elastic modulus/[GPa]</th>
<th>Yield strength/[MPa]</th>
<th>Strain hardening exponent/[$n$]</th>
</tr>
</thead>
<tbody>
<tr>
<td>SS 316L</td>
<td>227</td>
<td>210</td>
<td>290</td>
<td>0.35</td>
</tr>
<tr>
<td>As-cast HEA</td>
<td>150</td>
<td>244</td>
<td>160</td>
<td>0.77</td>
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<tr>
<td>388A-HEA</td>
<td>429</td>
<td>232</td>
<td>367</td>
<td>0.58</td>
</tr>
<tr>
<td>388C-HEA</td>
<td>625</td>
<td>211</td>
<td>507</td>
<td>0.61</td>
</tr>
</tbody>
</table>

3. Cavitation erosion behavior

Fig. 4 (a and b) shows the cumulative volume loss (CVL) and cumulative erosion rate (CER) as the function of time for the processed HEAs compared with as-cast HEA and stainless steel 316L. Further, the incubation period (IP) and mean depth erosion rate (MDER) for all the tested samples is shown in Fig. 4 (c and d). All the HEAs (unprocessed and processed) showed significantly higher cavitation erosion resistance than stainless steel 316L. The CVL of the SS 316L rises steeply after a short incubation period of 2.5 hours. No such sharp rise in the CVL was observed for the HEAs compared with stainless steel 316L (Fig. 4 (a)). The incubation period was also showed higher for all the HEAs. Among all, the highest incubation period was observed for 388C, which is around 4.2 times higher than stainless steel. As-cast HEA and 388A showed around 3 and 3.6 times better than stainless steel. The incubation period of 388C HEA showed nearly 1.2 and 1.6 times higher compared to 388A and as-received HEA. Higher the incubation period indicates plastic deformation with negligible damage. Furthermore, the deceleration stage preceded by a maximum erosion rate was observed (0.233 mm$^3$/h) after 12 h for stainless steel (Fig. 4 (b)). On the other hand, after higher incubation period, no acceleration-deceleration stage was observed for all the HEAs (Fig. 4 (b)). HEAs followed steady-state trend throughout the testing. After 20 hours of testing, the MDER of the stainless steel was around 1.46 µm/h followed by as-cast HEA (0.16 µm/h) (around 9 times erosion resistance). The 388C and 388A exhibited better cavitation erosion resistance compared to as-cast HEA with 20% and 16% lower erosion rate, respectively. The 388A and 388C showed around 15 and 20 times better erosion resistance than the stainless steel. The higher erosion resistance of the 388C is mainly attributed to the superior mechanical properties compared to other tested samples.

The correlation of different mechanical properties was evaluated to understanding the dominant mechanical properties controlling the cavitation erosion. The linear regression was performed to evaluate the correlation
between cavitation erosion resistance ($R_e$) and different mechanical properties. Cavitation erosion resistance is the reciprocal of mean depth erosion rate. Fig. 5 ((a to c)) showed cavitation erosion resistance ($R_e$) as the function of hardness ($H$), hardness to elastic modulus ($H/E$) and yield strength. The hardness showed a strong correlation (0.99) with cavitation erosion resistance (Fig. 5 (a)). Prior studies also showed a strong correlation for the hardness and yield strength with cavitation erosion resistance [45, 46]. The higher hardness and yield strength hinders the plastic deformation induced by micro-jets and high-intensity shockwaves. The higher cavitation erosion resistance observed for the 388C is due to the higher hardness and yield strength (Table 1) compared to other tested samples. The increased density of grain boundaries of processed HEAs act as slip barriers (small grain size compared to a coarser grain of as-cast HEA) by impeding dislocation movements which further restricts plastic deformation due to the impact of high velocity micro-jets and shockwaves. Furthermore, the presence of BCC phase (21% for 388C and 6% for 388A HEA) also limits the dislocation motions due to absence of preferential slip plane, which contributes to higher cavitation erosion resistance than the as-cast HEA. Recent study also showed the formation of bimodal (austenite and martensite) phase on stainless steel 316L helps in increasing the cavitation erosion resistance [47]. Furthermore, the combination of FCC and BCC phases favors the ductility and strength trade-off for the processed HEAs. Zhang, Ma, Zhao, Wu, Zhang, Wang and Qiao [48] also reported enhanced strength-ductility synergy for the NiCoCrFe HEA after deformation. The combined interactions also help in enhancing the cavitation erosion performance for the processed Al$_{0.1}$CoCrFeNi HEAs compared to as-cast HEA. Furthermore, the HEA are well known for having higher work hardening ability compared to traditional alloys owing to the lower stacking fault energy (SFE) [49]. Our previous study showed higher cavitation erosion performance for the as-cast HEA compared to stainless steel, despite the lower hardness (150 HV) attributed to higher work hardening ability [16]. As discussed previously, lower SFE is the strong function of number of elements, which contributes to increased work hardening and thereby better cavitation performance. The higher strain hardening exponent ($n$) was observed for all the HEA samples compared to SS316L (Table 1). The higher $n$ values for HEA indicate higher work hardening which results in higher cavitation erosion resistance compared to stainless steel. The higher $n$ value results in increased work hardening rate owing to the dislocation intersections. As a result, the flow stress increases which further restricts the plastic deformation and thereby improves the cavitation erosion resistance for HEAs compared to steel. The decrease in grain size (higher grain boundaries) lowers the strain hardening exponent ($n$) for the processed HEAs compared to unprocessed Al$_{0.1}$CoCrFeNi HEA due to restrictions for the movement of dislocations while increasing the yield strength due to grain-boundary strengthen. Prior studies
also reported similar behavior for different materials [44, 50]. The results showed that the effect of decreased work hardening (lower the $n$ value) has been circumvented through grain boundary strengthening for the processed HEAs. Significant grain refinement during FSP and the formation of BCC phases along the grain boundaries (388C) helps in enhancing the hardness and yield strength, which further augments to better cavitation erosion resistance for the processed HEAs compared to as-cast HEA. Besides hardness and yield strength, $H/E$ ratio showed a strong correlation, (adj.$R^2 > 90\%$) with $R_e$ (Fig. 5 (c)). The higher $H/E$ value indicates the material’s ability to undergo elastic deformation (lower plastic deformation). The processed HEAs showed higher $H/E$ value compared to as-cast HEA and stainless steel, which indicates resistance to higher resistance to plastic deformation and hence the cavitation erosion resistance. The cavitation erosion performance of the friction stir processed Al$_{0.1}$CoCrFeNi HEAs was compared to other conventional alloys and other HEAs [16, 47, 51, 52] are shown in Fig. 6. Interestingly, the 388A and 388C (bimodal) HEAs showed better cavitation erosion resistance than other alloys. The 388C HEA showed significantly lower erosion rate (3 times) than bimodal (austenite and martensite) stainless steel 316L developed using friction stir processing. The bimodal HEA also showed around 3 times higher erosion resistance compared to laser surface alloyed AlCoCrFeNi HEA. The significance of the HEAs is mainly attributed to the higher entropy effect resulting in severe lattice distortion and sluggish diffusion. The atomic size mismatch generally helps in elevating strain energy and dislocation pinning. The configurational entropy for Al$_{0.1}$CoCrFeNi is higher ($1.47R$) than that of the stainless steel ($1.16R$) resulted in sluggish diffusion which promotes the dislocation pinning capability. This helps in resisting the plastic deformation for the HEAs during the impact of microjets and shockwaves and thereby lowers the cavitation erosion rates compared to other conventional (processed and unprocessed) alloys.
Fig. 4. (a) Cumulative volume loss, (b) cumulative erosion rate, (c) incubation period and (d) mean depth erosion rate of as-cast and processed Al_{0.1}CoCrFeNi high entropy alloys compared with stainless steel 316L for erosion.
Fig. 5. Correlation parameters of cavitation erosion resistance of the unprocessed and processed Al₀.₁CoCrFeNi high entropy alloys with different mechanical properties (a) hardness, (b) yield strength and (c) hardness to elastic modulus.

Fig. 6. Comparison of mean depth erosion rates of friction stir processed Al₀.₁CoCrFeNi high entropy alloys with conventional alloys, friction stir processed stainless steel under different conditions and laser synthesized AlCoCrFeNi high entropy alloy [16, 47, 51, 52].

4. Damage morphology

The damage morphology of all the tested samples is shown in Fig. 7. The damage severity difference among the HEAs and stainless steel after 20 hours of testing are shown in Fig. 7 (a to d). It is evident from the macrograph that the steel showed deep pits and more profound damage compared to unprocessed and processed HEAs. No such kind of severe pits was observed for the HEAs. Scanning electron microscope showed striations along with deep pits and cracks are the major damage mechanism for stainless steel. When compared to stainless steel, the as-cast HEA showed shallow pits and micro-cracks (Fig. 7 (f and j)). Similarly, no such deep cracks and severe pits were present for the processed HEAs. Shallow pits and micro-cracks were also observed for both the processed samples. Furthermore, the presence of deep craters was observed for the steel which was not evident.
for HEA samples. The small-sized pits and cracks for all HEA samples might be related to its high work hardening behavior. Further, grain boundary strengthening also restricts the plastic deformation for both processed HEAs indicating lesser damage mechanism for the former. The Al₀.₁CoCrFeNi HEA have known for higher work hardening which has been reported earlier compares to conventional steels [16, 53]. During plastic strain, the dislocation interactions due to grain refinement further increase the flow stress contributing towards its better work hardening capability. Furthermore, the lower stacking fault energy (SFE) also increases the work hardening ability of the material resulting in an increase in the flow stress. This restricts the nucleation of microcracks and pits and thereby provide better cavitation erosion performance for the HEAs.

![Macrographs of tested samples](image)

**Fig. 7.** (a to d) Macrographs of all the tested samples under cavitation erosion condition for 20 hours. Scanning electron microscopy showing the low magnified images (e to h) and high magnified images (i to l) for all the tested samples.

5. Conclusion

In the present work, the friction stir processed Al₀.₁CoCrFeNi high entropy alloy was investigated under ambient and submerged cooling conditions at a constant rotational and transverse speed. Microstructural findings showed significant grain refinement after processing with equiaxed grains (≃ 1.6 µm) for 388A, whereas elongated grains (≃ 10 µm) surrounded by fine equiaxed grains (≃ 2 µm) for 388C highlighting bimodal structure. The hardness for the both 388C and 388A was around 625 HV and 429 HV, respectively which is
higher than that of as-cast HEA (150 HV). The hardness was around 4 and 3 times higher for both 388C and 388A compared to as-cast HEA. When compared to 388A, the hardness showed 1.6 times higher for 388C which is explained on the basis of higher fraction of BCC phases (21%) compared to 388A (6%). The increased hardness for 388C was mainly attributed to significant grain refinement and the presence of higher fraction of BCC precipitates. The cavitation erosion showed higher resistance for both processed HEAs (16 to 20%) which is explained based on significant grain refinement and the presence of BCC precipitates compared to single-phase FCC as-cast HEA. Morphology of the eroded samples showed deep craters and pits along with cracks for the stainless steel while HEAs showed striations with lesser number of pits and cracks. Detailed studies regarding the microstructure and the effect of the bimodal structure of HEA have considered being the prime scope of future work.

References


