Slurry erosion behavior of hydro-turbine steel treated cryogenically at different soaking periods

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Keywords: stainless steel-316, soaking period, cumulative mass loss

Abstract

The influence of deep cryogenic treatment on the erosive wear performance of Stainless Steel-316L (SS-316L) used in hydropower plants is studied. For this purpose, several SS-316L samples were held at deep cryogenic temperatures (−196 °C) for different soaking periods (12, 24, 36 h). The erosive wear tests were conducted on a self-fabricated slurry erosion test rig and the same was evaluated by weighing the cumulative mass loss (CML) of samples for every 30 min post erosion. From experimental analysis, it was found that the erosive wear was found to be minimum and the hardness reaches to maximum value after 24 h of the soaking period which could be attributed to the significant microstructural changes such as the transformation of γ-austenite phase into (δ-ferrite+α′- martensite) along with precipitation of numerous carbides after deep cryogenic treatments.

1. Introduction

Hydropower, as a renewable energy source, is vital for long-term electricity generation. However, slurry erosion deteriorates hydro-turbine components, which is a severe problem that many hydraulic power plants face worldwide [1]. Extremely hard and sharp abrasive particles entrapped in the flowing water strike on the surface of turbine components and alter the roughness and texture of the exposed surface, shortening their life. Slurry erosion depends upon various factors such as silt size and shape, hardness, concentration, angle of impingement, impact velocity etc [2]. It is not possible to find the common cause for slurry erosion [3]. In the recent past, many researchers tried to improve the erosion performance of hydro turbine steel by various heat treatment processes.

Many authors [3–21] reported that during the conventional heat treatment process (heating the material to austenitizing temperature, followed by quenching and tempering), all the austenite (fcc structure) could not be transformed into martensite (bct structure) and this untransformed austenite reduces the hardness and wear resistance of high-speed steels, austenitic steels, martensitic steel, and tool steels. So, for the complete transformation of this residual austenite into martensite, cryogenic treatment has been proposed. In cryogenic treatment, the material is submitted to sub-zero temperatures for a specific time followed by tempering. According to Cayron et al [22], during cryogenic treatment, the transformation of austenite to martensite is a very quick and diffusionless process that does not allow the carbon atoms to diffuse out and form carbides. The so formed martensite is very hard and exists in diverse forms such as lenticular, laths, thin plates, or butterfly.

Anna et al [23] claimed that the wear resistance is improved with deep cryogenic treatments (DCT) because of the transformation of residual austenite into martensite and homogenous distribution of carbides. Zhirafar et al [24] studied the effect of DCT on the mechanical properties of 4340 steel and found that the hardness and fatigue strength of cryogenically treated samples was more than conventionally heat-treated samples. Upadhyay et al [25] investigated the mechanical and metallurgical properties of SS316 after deep cryogenic treatment.
Tempering was done at two temperatures: 350 °C and 250 °C, after the deep cryogenic treatment. The samples that were tempered at 250 °C had higher hardness and tensile strength, according to the findings. Baldissera et al [26] studied DCT on AISI 302 Stainless-steel and concluded that DCT can improve the fatigue life of the AISI 302 stainless steel while there is no effect of DCT on its corrosion resistance. Sugavaneswaran et al studied the effect of DCT on the wear behavior of SS316L steel and found a noticeable improvement of 28.54% in its microhardness after DCT. Smallman et al [27] reported that the formation of martensite from austenite takes place below 200 °C and this transformation is due to the lattice distortion or shear mechanism in the matrix. According to Bidulsky et al [28], cryogenic treatment is the procedure for removing retained austenite from carbon alloyed steels which necessitates lowering the temperature below zero degree Celsius for the transformation of retained austenite to martensite. Ghosh et al [29] reported the presence of martensite phase with grain boundary precipitates in the sensitised austenitic stainless steel. Hedstrom et al [30] claimed that the high alloyed austenitic stainless steels are more stable, and will only transform to martensite at cryogenic temperatures. The author also explained the diffusionless transformation of austenite (fcc) into α-martensite (bct) in accordance with Bains’s theory. Singh et al [31] reported that heat treatment is the only economical alternative to improve tribological properties of SS316L. Javid et al [32] reported that the essential chemical driving force for the initiation of martensitic transformation is the difference in free energies between austenite
and martensite at martensitic start temperatures. Nalepka et al [33] reported that the change of face-centred cubic (fcc) austenite into body-centred cubic (bcc) martensite is accompanied by a significant rise in yield stress when the temperature is reduced to near absolute zero. Hassan et al [34] studied erosive wear performance of 13Cr-4Ni martensitic stainless steel heat treated at austenitizing temperatures of 950 °C, 1000 °C and 1050 °C for 2, 4 and 6 h soaking periods followed by oil quenched and tempering for 1 h at a 600 °C. The author reported that after heat treatment, the weight loss in steel was approximately 34% less as compared to as-received cast steel. According to Das et al [35], the DCT can be used to process a broad range of materials, along with ferrous and non-ferrous materials, alloy steels, copper alloys, and plastic materials. Hull et al [36] studied delta ferrite and martensite formation in stainless steels. According to the author, when an austenitized alloy with 14 per cent Chromium and 14 per cent Nickel was cooled to −320 F in liquid nitrogen and warmed to room temperature, 13 per cent martensite was formed. Zhi et al [37] studied the effects of cryogenic treatment on mechanical properties and microstructure of Fe–Cr–Mo–Ni–C–Co alloy. It was found that after cryogenic treatment, the originally retained austenite was separated into small pieces by the needle-like martensite. The hardness of the alloy is increased after cryogenic treatment, which is mainly attributed to the transformation of the retained austenite into martensite. According to Hedstrom et al [38], austenitic stainless steels consists of two types of alloying elements i.e., ferrite and austenite stabilizers, with Cr being a ferrite stabilizer and Ni being an austenite stabilizer. In these metastable stainless steels, austenite can change into two forms of martensite, one of which is hexagonal close-packed (HCP) martensite, also known as ε-martensite. The other is α′-martensite, which is a body-centred cubic (BCC) martensite. Tian et al [39] studied martensite (ε and α′) formation during incremental cooling of Fe–Cr–Ni alloys. It was found that as the temperature decreases, the volume fraction of α′-martensite increases. Patil et al [19] reviewed the effects of cryogenic treatment on different types of steels and concluded that Cryogenic treatment (CT) is the supplementary process to conventional heat treatment process in steels, and tends to enhance its mechanical and physical properties.

During cryogenic treatments, the soaking period is crucial for structural transformation from austenite to martensite, carbide formation, and carbide distribution. In cryogenic treatment, optimising the soaking period is critical for improving efficiency and lowering costs. One group of researchers [11, 40, 41] claims that the optimum soaking period is 24 h, while another group [3, 42] claims 36 has optimum soaking period. Some researchers [43, 44] also claims 48 h soaking period as optimum. The concepts of the optimum soaking period are not fully understood in the literature. However, some researchers have found that soaking for shorter or longer periods of time yielded greater results. The results for various soaking periods differed depending on the base material. There is no appropriate explanation for the soaking phenomenon in the cryogenic treatment process. Further, most of the literature findings on cryogenic treatments are limited only to tool steels. Thus, there remains an open question that whether the same treatment could be effective to improve the wear resistance of SS-316L used in hydropower plants. To prove this hypothesis, the group of SS-316L samples were

| Element composition at 24 °C | C | Si | Mn | P | S | Ni | Cr | Mo | Fe |
|----------------------------|---|----|----|---|---|----|----|----|----|
| Percentage                 | .02| .37| 1.35| .039| .007| 10.20| 16.12| 2.06| 69.83|

Table 1. The elemental composition of SS-316L.

and martensite at martensitic start temperatures. Nalepka et al [33] reported that the change of face-centred cubic (fcc) austenite into body-centred cubic (bcc) martensite is accompanied by a significant rise in yield stress when the temperature is reduced to near absolute zero. Hassan et al [34] studied erosive wear performance of 13Cr-4Ni martensitic stainless steel heat treated at austenitizing temperatures of 950 °C, 1000 °C and 1050 °C for 2, 4 and 6 h soaking periods followed by oil quenched and tempering for 1 h at a 600 °C. The author reported that after heat treatment, the weight loss in steel was approximately 34% less as compared to as-received cast steel. According to Das et al [35], the DCT can be used to process a broad range of materials, along with ferrous and non-ferrous materials, alloy steels, copper alloys, and plastic materials. Hull et al [36] studied delta ferrite and martensite formation in stainless steels. According to the author, when an austenitized alloy with 14 per cent Chromium and 14 per cent Nickel was cooled to −320 F in liquid nitrogen and warmed to room temperature, 13 per cent martensite was formed. Zhi et al [37] studied the effects of cryogenic treatment on mechanical properties and microstructure of Fe–Cr–Mo–Ni–C–Co alloy. It was found that after cryogenic treatment, the originally retained austenite was separated into small pieces by the needle-like martensite. The hardness of the alloy is increased after cryogenic treatment, which is mainly attributed to the transformation of the retained austenite into martensite. According to Hedstrom et al [38], austenitic stainless steels consists of two types of alloying elements i.e., ferrite and austenite stabilizers, with Cr being a ferrite stabilizer and Ni being an austenite stabilizer. In these metastable stainless steels, austenite can change into two forms of martensite, one of which is hexagonal close-packed (HCP) martensite, also known as ε-martensite. The other is α′-martensite, which is a body-centred cubic (BCC) martensite. Tian et al [39] studied martensite (ε and α′) formation during incremental cooling of Fe–Cr–Ni alloys. It was found that as the temperature decreases, the volume fraction of α′-martensite increases. Patil et al [19] reviewed the effects of cryogenic treatment on different types of steels and concluded that Cryogenic treatment (CT) is the supplementary process to conventional heat treatment process in steels, and tends to enhance its mechanical and physical properties.

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cryo-treated at −196 °C (with distinct holding periods of 12 h, 24 h and 36 h). Further, tests were conducted on the slurry test rig to analyse the erosion wear performance of SS-316L samples and to find the optimum soaking period to maximize its wear resistance. The complete research methodology of the present study is presented in figure 1.

2. Experimentation

2.1. Sample preparation
The samples were cut from commercially available AISI 316 L stainless steel to a size of 25 × 25 × 5 mm. The nominal chemical composition of the base material is shown in table 1 and chemical composition according to ASTM standard is shown in table 2. AISI 316 L has wide applications in hydro-power plants for the fabrication of various parts such as impellers of submersible pumps [45, 46]. After cutting, the samples were polished as per the ASTM standard E3–11 to reveal the metallographic changes [42].

2.2. Conventional and Deep cryogenic treatments
The samples were submitted to Conventional heat treatment (CHT) and Deep-cryogenic treatment (DCT) in separate batches. The CHT comprises of austenitizing, quenching and dual tempering (T), i.e., pre and post

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Table 2. The elemental composition is according to ASTM standards.

| Element composition at 24°C | C  | Si | Mn | P  | S   | Ni | Cr   | Mo | Fe   |
|-----------------------------|----|----|----|----|-----|----|------|----|------|
| Percentage                  | .03| .75| 2  | .045| .03 | 10–14| 16–18| 2–3| 69.83 |

Table 3. Four distinct heat-treatments of SS-316L.

| Sample code | Hardening | Pre-tempering | Cryogenic treatment | Post-tempering |
|-------------|-----------|----------------|---------------------|----------------|
| CHT         | 1 h at 920°C followed by oil-quenching | 2 h at 200°C | None | 2 h at 150°C |
| DCT12       | —         | —              | 12 h at −196°C      | 2 h at 150°C   |
| DCT24       | —         | —              | 24 h at −196°C      | 2 h at 150°C   |
| DCT36       | —         | —              | 36 h at −196°C      | 2 h at 150°C   |
tempering, which was according to the ASM Heat Treater’s Guide [47]. The distinct heat-treatments performed on SS316L are presented in table 3.

The cryogenic treatment was carried out in the computer-controlled cryogenic equipment comprising of an insulated cryogenic compartment, thermocouple, temperature regulator and nitrogen tank available at the Institute of Auto parts, Ludhiana (India) as shown in figure 2. Nitrogen gas was used as a cooling medium. The deep cryogenic treatment was performed by slow cooling of the samples to −196 °C, and holding the samples at this temperature for specific time frames (12, 24, 36 h), followed by tempering at 150 °C for 2 h. Also, a schematic diagram of the complete heat treatment process with heating and cooling conditions is presented in figure 3.

2.3. Microstructural characterization
The presence of different phases such as γ-austenite, δ-ferrite and α′-martensite was analysed using X-ray diffraction (PANalytical X’Pert PROMPD) with Cu-K radiation between 30°–90° Bragg’s angle at 40 kV, in accordance with ASTM E975-00 [42]. The microstructural analysis of the SS-316L specimens prepared in section 2.1 was performed at IIT, Roorkee (India). For a detailed metallurgical study, the sample’s surface was detected by Scanning electron microscope (SEM, Carl Zeiss EVO 40 XVP) after etching in the 122 ml HCL, 6 ml HNO3, 8.5 gm FeCl3, 2.4 gm CuCl2, 122 ml C2H6O etchant for 10–15 seconds. The carbide content was calculated by using image analyzing software (Clemex Vision, version 3.5.025).
Table 4. Preparation of 300 μm average sand size.

| B.S.S no | Mesh size (μm) | Multiplier (Mi) | Weight used in gm (Wt.) | Mi X Wi |
|----------|---------------|----------------|-------------------------|---------|
| 52       | 300           | 36             | 1500                    | 54000   |
| 30       | 250           | 52             | 2500                    | 130000  |
| 72       | 212           | 30             | 1000                    | 30000   |
| 100      | 150           | 72             | 300                     | 43200   |
| 150      | 106           | 100            | 300                     | 30000   |

Total weight = 6200 gm, Total = 347200

Average finest number of sand grains, AFN = 347200/6200
AFN = 56; AFN = B.S. S no
British standard specifications, B.S.S no = 56 ≈ 300 μm
2.4. Macro-hardness

The macro-hardness (HRC) was calculated using a Rockwell hardness tester with a load of 1.5 kN as per ASTM E18.

2.5. Collection and preparation of slurry for erosion testing

The silt particles were collected from Bhakra sub-power station situated at Bilaspur in Himachal Pradesh, India. Further, the silt particles were dried and sieved to separate different sizes. Finally, these sand grains were mixed in adequate ratio to get desired average size of 300 μm. Bansal et al [46] also used similar parametric values to study the slurry erosion performance of SS-316 Steel coated with HVOF WC-10Co-4Cr identical to our study. The particle size distribution of the silt particles is shown in figure 4. Table 4 shows the sand preparation with an average size of 300 μm. The morphology of silt particles is uneven with sharp edges as presented in figure 5. The sand was then mixed in distilled water to form a slurry with a concentration of 20,000 ppm. A similar concentration of slurry was used by Singh et al [48] to study the Slurry erosion performance of HVOF Sprayed Coatings on Hydraulic Turbine Steel identical to our study.

2.6. Fabrication of slurry erosion test rig

A slurry erosion testing apparatus was self-fabricated in accordance with ASTM standard G-73 to perform slurry erosion experiments, as shown in figure 6 [49, 50]. Post experimentation, the eroded sample was cleaned with ethanol and mass loss of each worn-out sample was weighted for every 30 min. The overall duration of each experiment was 210 min. The erosive wear was calculated from the cumulative mass loss of each sample.

As presented in table 5, the most significant parameters such as impact velocity, impingement angle with two different levels (min. and max.) opted for experimentation are in accord with those considered by authors [51–54] to study the slurry erosion behavior of steels. Grewal et al [55] in one of their studies also conducted the slurry erosion experiments at 16 m s−1 impact velocity and impact angle between 30° to 90° identical to our experimental conditions. Vyas et al [45] also conducted slurry erosion experiments of SS316 at 30° and 90° and impact velocity of 14.7 m s−1 to 30.6 m s−1 in accord with our study. Table 6 presents the Taguchi L4 experimental design for conducting experimental work. Each experiment was performed five times and the mean values were considered for the erosion results.

| Experiment no | Impingement angle in degrees (A) | Impact vel. of particles in m/s (B) |
|---------------|----------------------------------|-----------------------------------|
| E1            | 90                               | 10.04                             |
| E2            | 90                               | 14.32                             |
| E3            | 30                               | 10.04                             |
| E4            | 30                               | 14.32                             |

Table 5. Slurry erosion parameters at different levels.

Table 6. Taguchi L4 experimental design.

3. Results and discussion

3.1. Microstructural analysis

The surface morphology of CHT samples indicates the presence of γ-austenite and skeletal type δ-ferrite (figure 7(a)). The high peaks of Iron, Nickel and Manganese in EDS analysis also validate the presence of a large amount of γ-austenite in the CHT samples (figure 7(a)′). Since the austenitic stainless steels are not stable at room temperature and get transformed into γ-austenite, δ-ferrite, and carbides [56]. Therefore, our finding is in good agreement with this context. After DCT 12 h, the skeletal structure of ferrite shrinks and dendrites of δ-ferrite were formed with small laths of martensite in the austenitic matrix (figure 7(b)). The possible reason could be the rapid cooling of SS-316L at ultra-low temperatures. At these temperatures, the austenite gets transformed into martensite and further upon tempering again decomposes into ferrite with a small amount of
martensite present with it. The decrease in the levels of austenite forming elements such as Nickel and Manganese in EDS analysis also suggests the diminishing of the austenite phase (figure 7(b’)).

After DCT 24 h, the microstructure gets refined and there was a formation of uniform δ-ferrite dendrites in the austenitic matrix along with some signs of martensitic laths as shown in figure 7(c). Also, the precipitation of homogenously distributed fine carbides can be seen in the microstructure of DCT 24 h samples, which could be because, at low temperatures (−196 °C), the structure of the SS-316L undergoes an excessive contraction (lattice distortion) and causes the generation of high dislocation density in the matrix-carbide interface which acts as nuclei for the precipitation of new carbides during tempering. The presence of high peaks of Iron and Chromium emplace of high peaks of Nickel and Manganese in EDS analysis also suggests the presence of martensite (figure 7(c’)).

After DCT 36 h, the microstructure of SS-316L depicts the presence of δ-ferrite dendrites, martensitic laths, and carbides. However, the number of small carbides decreases and their homogeneity breaks (figure 7(d)). This could be due to the growth of the carbon clusters that increase the size of carbide precipitates, thereby decreasing the number of the carbides and their homogenized distribution. Ibrahim et al [42] also proposed a similar theory for the decrease in the number of carbides identical to our result. The presence of high peaks of Iron and Chromium in EDS analysis also suggests the formation of martensite and carbides (figure 7(d’)).

The authors Weng et al [57], Singh et al [58], Li et al [59], Chen et al [60], Wu et al [20], found similar findings during cryogenic treatments of different alloys steels. Also, the author Li et al [59], claimed an 8.9% rise in volume-percentage of carbides after D-CT. Amini et al [44, 61] observed a 6% rise in the vol.-percentage of small-carbides after D-CT. The author Podgornik et al [62] also performed D-CT on cold worked tool steel and reported the formation of a needle-like martensitic structure in addition to the formation of small carbides after D-CT. In one of their studies, Gill et al [3] reported 34.02 per cent escalation of the small carbides after D-CT respectively.

The immensity of carbide was calculated with the help of optical micrographs using image analyzing software (Clemex-Vision @ version 3.5.025). A similar technique was used by a number of authors [41, 43, 63, 64] to calculate the volume percentage of carbides. The carbides of size greater than 4 μm are termed as large carbides while sizes less than 4 μm are termed as small carbides. A similar concept of small and large carbides have been proposed by authors [12, 65–67]. It was estimated that the small carbide immensity increases by 6.5%(vol) and 30.4%(vol) after DCT 12 h and DCT 24 h respectively as compared to CHT. However, the immensity of small carbides decreases during DCT 36 has compared to DCT 24 h suggesting that the optimum soaking period for saturation of carbide precipitation is 24 h. This concept of the longer soaking period is in accord with the studies conducted by authors [42, 65, 68].

3.2. X-Ray diffraction analysis

Figure 7 reveals the X-Ray diffraction analysis of the CHT, DCT 12 h, DCT 24 h and DCT 36 h samples. The key advantage of XRD is its ability to detect and estimate the different phases such as γ-austenite, δ-ferrite and α’-martensitic by analysing their crystal structures that could not be detected by light optical microscopy or
scanning electron microscopy [41]. The peaks analogous to planes (110$\delta$), (202$\alpha'$) corresponds to $\delta$-ferrite and $\alpha'$-martensite phase respectively with BCC structure whereas, the peaks analogous to planes (111$\gamma$), (200$\gamma$), (220$\gamma$) and (311$\gamma$) corresponds to austenite with FCC structure (figure 8). The XRD spectrum of untreated

Table 7. Micro-hardness values (HV) of CHT, DCT 12 h, DCT 24 h and DCT 36 h samples.

| Treatment   | Macrohardness (HRC) |
|-------------|----------------------|
| CHT         | 24                   |
| DCT 12 h    | 29                   |
| DCT 24 h    | 33                   |
| DCT 36 h    | 31                   |

Figure 9. The graph between (a) cumulative mass loss versus time, (b) wear rate versus time (for experiment no-1).
samples consists of high \((111\gamma)\) peaks suggesting that austenite is the main phase in the SS-316L. However, minute traces of \(\alpha'\)-martensite were also detected in the untreated samples in the form of small humps associated with the \((202\alpha')\) peak (figure 3). These minute amounts of ferrite or martensite could be produced during manufacturing processes by heavy thermomechanical processes. After DCT 12 h, the intensity of peak \((200\gamma)\) and \((311\gamma)\) decreases suggesting the transformation of austenite into \(\delta\)-ferrite or martensite. After DCT 24 h, the intensity of peak analogous to the plane \((200\gamma)\) and \((311\gamma)\) further reduced to a very low level, implying the transformation of residual austenite into martensite. Also, there was a precipitation of fine carbides \((\text{Cr}_{26}\text{C}_6)\) after DCT 24 has shown in figure 7. Further, after DCT 36 h, there was not much difference in the intensity of peaks implying the threshold limit of phase change in SS-316L.

### 3.3. Macro hardness

The increase in the microhardness was found to be about 20.8%, 37.5% and 29.1% after DCT 12 h, DCT 24 h and DCT 36 h when compared to the CHT sample. The phase transformation from retained austenite to martensite and the nucleation of carbide particles in the microstructure is responsible for the increase in hardness value. Kumar et al and Soleimany et al\(^{[9, 41]}\) also reported similar causes for the increase in hardness. However, as the holding duration increases from 24 to 36 h, the microhardness decreases significantly, as seen in
The decline in microhardness could be attributed to carbide particle disintegration in the microstructure. Ibrahim et al.\cite{42} reported similar results identical to our studies.

### 3.4. Erosive wear performance

The erosion wear performance of the CHT, DCT 12 h, DCT 24 h, and DCT 36 h steel samples were investigated using a slurry erosion test apparatus (jet type) under various experimental conditions. The Cumulative Mass Loss (CML) trend for differently heat-treated samples indicates that initially, the CML was almost linear implying that the erosion remains constant, thereafter there was a slight decrease in the CML (figure 9(a)–12(a)). This could be attributed to strain hardening of the targeted material after a number of impacts with erodents. Grewal et al.\cite{53} in one of their studies also discussed the concept of strain hardening identical to our study.

Further, the CML was found maximum at low impingement angles and high impact velocities which could be because of the increase in kinetic energy of erodents at high velocities. Also, at low impact angles the high CML is attributed to the cutting action of the erodents striking the targeted surface tangentially, whereas, at high impact angles, the low CML is attributed to the fact that when the erodents strike the targeted surface at 90°, the eroded material could not get sufficient time to escape from the eroded surface and may act as a barrier between the targeted surface and the fresh incoming erodents which eventually reduces the erosion wear rate (figure 13).

![Figure 11. The graph between (a) cumulative mass loss versus time, (b) wear rate versus time (for experiment no–3).](image-url)
A similar concept of plastic deformation at 90° of impingement angles have been proposed by Sharma et al [69], Burstein et al [70] and Nandre et al [71] identical to our study. Furthermore, the minimum CML was reported for DCT 24 h samples, while the maximum CML was for CHT samples for all experimental conditions. The least CML in DCT 24 h samples could be attributed to its tremendous hardness. Similar results have been proposed by Amini et al and Soleimany et al [41, 44] in their studies.

The graph between the Wear rate trend (figures 9(b)–12(b)) indicates that initially the wear rate increases abruptly for 100 min. Afterwards, there was a fall in the wear rate and then tries to remain constant which could be attributed to the work hardening phenomenon during the continuous impact of erodents on the targeted surface. In all the experimental conditions, the CML and wear rate was found to be maximum for CHT samples and minimum for DCT 24 h samples. Low wear rate could be attributed to the various microstructural changes resulting after deep cryogenic treatment for 24 h of soaking period i.e., the transformation of residual austenite into martensite, refinement of microstructure and precipitation of fine carbides.
3.5. SEM Analysis of eroded samples

The process of erosion wear is greatly affected by the properties of the targeted surface, the shape of the erodent, impingement-angle and impact velocity [69, 72]. It was observed that at 90° of impingement angle, the removal of material was mainly because of two mechanisms i.e., formation of craters and plastic deformation (figure 13(a)). The surface morphology of eroded CHT samples reveals the presence of deep crater marks with the formation of platelets (figure 14(a)). However, the crater marks are shallow in the case of DCT 12 h, DCT 24 h and DCT 36 h eroded samples which could be due to the tremendous hardness of the deep cryogenically treated samples (figures 14(b)–(d)). It is interesting to note that the formation of craters was less severe in eroded DCT 24 h samples as compared to DCT 12 h and DCT 36 h samples. The possible reason could be the presence of hard carbides in DCT 24 h samples that resist the huge impacts of rodents. Further, signs of plastic deformation can also be seen at various locations due to a number of impacts by the erodents. Grewal et al [2] also proposed that at normal impingement angles material removal takes place in the form of platelets and plastically deformation.

At low rake angles, the normal component of the impact force imparted by the impacting particles to the target material is small. Therefore at 30° of impingement angle, the wear marks caused by the striking erodents appear as shallow scars on the skin of the targeted material (figure 13(b)). In the light of this argument, scars of

Figure 13. Schematic diagram showing the slurry erosion mechanism at an impingement angle (a) 90-degree (b) 30-degree.
wear along with ploughing and cutting marks can be seen in the fractography of eroded CHT, DCT12h, DCT24h and DCT36h samples (figures 15(a)–(d)). But there are certain differences in the wear mechanism of all considered samples. The eroded CHT samples reveal huge marks of ploughing and cutting due to unrefined microstructure. Whereas, the severity of the ploughing mechanism is less in eroded DCT 12 h samples and the least severity of ploughing was reported in DCT 24 h samples followed by DCT 36 h samples which could be due to their tremendous hardness and refined microstructure. However, the intensity of ploughing marks was recorded less in DCT12h, DCT24h and DCT36h samples (figures 15(b)–(d)). These findings of maximum erosion at low impingement angles are in agreement with the facts reported by authors [55, 73–75].

4. Conclusions

The objective of this study was to prove the hypothesis that whether the cryogenic treatment could be effective to improve the wear resistance of SS-316L used in hydropower plants and to find the optimum soaking period of cryogenic treatment to improve its wear resistance. Following observations have been made from the results of the study:

1. The small martensitic laths along with δ-ferrite dendrites were formed in the austenitic matrix after deep cryogenic treatments for different soaking periods (DCT 12 h, DCT 24 h, DCT 36 h) which could be because of the diffusionless transformation of residual austenite into martensite at very low temperatures.

2. The immensity of small carbides (of size 0–4 μm) uniformly distributed throughout the matrix has been increased after deep cryogenic treatment for 24 h of the soaking period, while decreased during 36 h of soaking period suggesting that the optimum soaking period for saturation of carbide precipitation is 24 h.

3. It was found that the macro-hardness has been increased to 29HRC, 33HRC, 31HRC after deep cryogenic treatments with a soaking time of 12 h, 24 h, 36 h respectively in comparison with 29HRC of the conventionally heat-treated sample. The possible reason could precipitation of small carbides after deep cryogenic treatments.
4. The erosive wear performance of cryogenically treated samples (with a soaking period of 24 h) was found maximum. However, further, an increase in the soaking period results in decreased erosive wear resistance. This could be attributed to that during long soaking periods of more than 24 h, the segregation of atoms results in the extreme growth of the precipitated carbides. As a result, the carbide distribution weakened which further reduces the wear resistance.

5. The cumulative mass loss was reported maximum at 30° of impingement angle and minimum at 90°, suggesting the ductile behavior of SS-316L material.

6. It was observed that at 90° of impingement angle, the removal of material was mainly because of two mechanisms i.e., formation of craters and plastic deformation. However, at a 30° impingement angle, the material removal is primarily due to ploughing and cutting action.

Data availability statement
All data that support the findings of this study are included within the article (and any supplementary files).

Declaration of conflicting interest
The authors declared no potential conflicts of interest with respect to the research, authorship, and publication of this article.

Funding
The authors received no financial support for the research, Authorship, and/or publication of this article.

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