Enhancing the cavitation erosion resistance of AISI 420-type stainless steel with quenching and partitioning

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ABSTRACT

Stainless steels are commonly used in hydraulic components where they may be susceptible to cavitation erosion. In this study the cavitation erosion resistance of AISI 420-type stainless steel is examined after quenching and partitioning (Q&P) heat treatment. Q&P-samples were prepared with varying heat treatment parameters, and their initial properties were examined with X-ray diffraction and hardness measurements. Reference samples were also prepared with quenching and tempering and quenching without partitioning. The samples were eroded for 6 h with an ultrasonic cavitation erosion device, and their mass losses were measured. The eroded areas were examined with scanning electron microscopy, optical profilometry, and magnetic induction measurements. The results suggest that the cavitation erosion resistance of the examined stainless steel can be significantly enhanced with Q&P. This enhancement of cavitation erosion resistance results from high initial hardness and retained austenite fraction of the steel. During cavitation erosion the retained austenite can absorb cavitation bubble collapse energy by transforming into strain induced martensite, which increases the hardness of the steel and generates expansion of the lattice. This expansion additionally hinders crack propagation at grain boundaries, which reduces the formation of initial cavitation damage.

1. Introduction

Stainless steels are common material choice for several hydraulic components such as valves, pumps, pipes, turbines, and compressor blades [1,2]. During the operation these components are often subjected to cavitation erosion, which leads to degradation of the component surface over time. The hydraulic profile of the surface has a significant influence on the efficiency of hydraulic equipment. Therefore, enhancing the cavitation erosion resistance of the selected stainless steels can reduce maintenance costs and increase system efficiency over time [1]. As a result, there have been many studies on the improvement of cavitation erosion resistance covering a wide range of stainless steel alloys [1,3–8]. It has been observed [3,6,9] that increasing hardness of the material results in better resistance to cavitation erosion. Hence, hardening phenomena related to plastic deformation during cavitation erosion, such as grain refinement [6,10] and strain induced martensitic transformation [1,11], may have a significant effect on enhancing the cavitation erosion resistance. These observations suggest that the cavitation erosion resistance of stainless steels could be improved by tailoring their microstructure with novel heat treatments which promote strain induced hardening. One such method could be quenching and partitioning (Q&P).

Quenching and partitioning, as originally proposed by Speer et al. [12], is a heat treatment in which steel is first austenitized and then quenched to a temperature between the martensite start (M_s) and finish (M_f) temperatures, followed by a partitioning treatment at elevated temperature and final quenching to room temperature. The aim of Q&P is to form a fine microstructure comprising martensitic matrix and retained austenite, which would result in high strength of the material without sacrificing its ductility. As Q&P is applicable for stainless steels, there has been many studies [13–19] focusing on the mechanical properties of quenched and partitioned stainless steels. These studies suggest that transformation-induced plasticity (TRIP) effect is prominent and contributes to the strain hardening in stainless steels after Q&P. Considering that strain induced martensitic transformation has been observed to enhance resistance to cavitation erosion in several studies [11,20–24], Q&P could be used for applications where cavitation occurs as well. To the authors knowledge, however, little research has been made on the cavitation erosion behaviour of stainless steels after Q&P. Earlier study by Zhou et al. [25] showed that Q&P could be used for
improving the cavitation erosion resistance, as the cavitation-induced martensitic transformation of the retained austenite increases the hardness of the steel. Although their study examined an ultra-low carbon CrMnN stainless steel, the same principles should apply for other quenched and partitioned stainless steel grades as well. For example, AISI 420-type stainless steels, which have been investigated in different cavitation erosion studies [2,26], are also applicable for Q&P heat treatment [14,15,17]. Therefore, quenched and partitioned AISI 420-type stainless steels could be promising candidates for applications susceptible to cavitation erosion.

The purpose of this study is to investigate the cavitation erosion resistance of AISI 420-type stainless steel after Q&P heat treatments. The steel samples were quenched and partitioned using varying temperatures and times and subjected to ultrasonic cavitation erosion. Reference samples in as-quenched, and quenched and tempered states were additionally made for comparison. The cumulative mass losses were measured for all samples, and the eroded surfaces were characterized with electron microscopy and optical profilometry. Additionally, the microstructural changes after cavitation erosion were examined with magnetic induction measurements.

2. Experimental procedures

The steel chosen for this study was an AISI 420-type stainless steel grade EN 1.4034 with chemical composition shown in Table 1. The steel was delivered as hot rolled and pickled (surface finish 1D [27]) sheet with dimensions of 6.0 mm × 1250.0 mm × 2000.0 mm. Samples for the cavitation tests were cut in a rectangular shape with rounded corners with width and height of 20.0 mm × 20.0 mm.

The heat treatment parameters were chosen based on the authors’ earlier work [17], common recommendations for the quenching and tempering of the steel [27], as well as a combination of these two. An illustrative diagram of the used heat treatments is presented in Fig. 1. All samples were austenitized at 1100 °C for 15 min, followed by quenching in oil for 10 s. The quenching temperature was varied between 50 °C and 170 °C for the Q&P samples, whereas other samples were quenched to room temperature (RT). The samples were then heated to varying tempering or partitioning temperatures, excluding one sample that was left as reference after the initial quenching to room temperature. For samples partitioned at 400 °C a salt bath was used, whereas other samples were heated to 200 °C in air. After tempering or partitioning the samples were water-quenched to room temperature. The samples were named according to the applied heat treatment: directly after quenching as DQ, quenched and tempered sample as Q&T, and Q&P samples based on used temperatures as seen in Table 2. The name of the combination sample Q50P200 includes the partition temperature, as it differs from the 400 °C of other Q&P samples.

The selection of chosen heat treatment parameters was validated with dilatometry using a Bähr DIL 805 A/D dilatometer. The sample for dilatometry was in cylindrical shape with length of 10 mm and diameter of 4 mm. The sample was austenitized by heating it to 1100 °C with a heating rate of 20 K/s. The holding time was 15 min, after which the sample was quenched with a rate of 20 K/s to room temperature. The austenite start (A s), austenite finish (A f), and M t temperatures of the steel were determined based on the corresponding length changes.

After the heat treatments the sample surfaces were polished for cavitation tests. The grinding stage was performed using SiC-papers with decreasing grain size. The samples were then polished with cloths using 3 μm and 1 μm diamond suspensions and finished to abrasive size of 0.02 μm with MasterMet 2 suspension. The initial hardness of the samples was measured using a Struers Duramin-A300 hardness testing system. The measurements were performed according to standard EN ISO 6507–1:2018 [28] as low-force Vickers hardness tests with a load of 2 kg (HV2).

The cavitation erosion tests were carried out according to standard ASTM G32-16 [29] using a Sonics & Materials Inc. VCX 750 ultrasonic vibrator. The used power and frequency were 750 W and 20 kHz, respectively. The material of the tip was Ti-6Al–4V and the tip diameter was 15.9 mm. The samples were submerged in deionized water, and the test temperature was kept at 25 ± 2°C. The samples were placed at a distance of 0.5 mm parallel to the vibrating tip. The mass loss of the samples was measured after test times of 15 min, 30 min, 1 h, 2 h, 4 h, and 6 h. After every interval the samples were ultrasonically cleaned in ethanol, dried, and weighed with a laboratory scale. The vibrating tip was polished every 2 h to remove wear damage from the tip.

After the cavitation erosion tests the samples were further analysed with electron microscopy and optical profilometry. Microstructural characterization of the cavitation-eroded surfaces was conducted using a Zeiss ULTRApplus field emission scanning electron microscope (FE-SEM). The used acceleration voltage was 15 kV, and the working distance was set to approximately 8.5 mm. The features of the eroded surfaces were further examined with Alicona InfiniteFocus G5 optical 3D profilometer.

The initial volume fraction of retained austenite before cavitation tests was estimated by measurements with X-ray diffraction (XRD). The XRD measurements were carried out with a Stresstech Xstress 3000 G X-ray diffractometer. The measurements were performed following the ASTM standard E957 – 13 [30] by using CrKα-radiation with a collimator size of 3 mm. The used voltage and current values were set at 30 kV and 6.7 mA, respectively. The Kα-radiation was removed by using special filters on the detectors. The detectors were set to gather 20-peaks at 156.4° (211) and 106.1° (200) with an exposure time of 10 s for ferrite

Table 1

<table>
<thead>
<tr>
<th>Steel grade</th>
<th>C</th>
<th>Cr</th>
<th>Mn</th>
<th>Si</th>
<th>P</th>
<th>S</th>
<th>Fe</th>
</tr>
</thead>
<tbody>
<tr>
<td>EN 1.4034</td>
<td>0.45</td>
<td>13.7</td>
<td>0.55</td>
<td>0.35</td>
<td>0.029</td>
<td>0.002</td>
<td>Bal.</td>
</tr>
</tbody>
</table>

Fig. 1. Illustration of the used heat treatment cycles. Tempering is highlighted with red color. (For interpretation of the references to color in this figure legend, the reader is referred to the Web version of this article.)

Table 2

<table>
<thead>
<tr>
<th>Sample</th>
<th>Austenitization</th>
<th>Initial quench</th>
<th>Partitioning/ Tempering</th>
<th>Final quench</th>
</tr>
</thead>
<tbody>
<tr>
<td>DQ</td>
<td>1100 °C/15min</td>
<td>RT</td>
<td>–</td>
<td>–</td>
</tr>
<tr>
<td>Q&amp;T</td>
<td>1100 °C/15min</td>
<td>RT</td>
<td>200 °C/2 h</td>
<td>RT</td>
</tr>
<tr>
<td>Q50P200</td>
<td>1100 °C/15min</td>
<td>50 °C</td>
<td>200 °C/1 h</td>
<td>RT</td>
</tr>
<tr>
<td>Q60P</td>
<td>1100 °C/15min</td>
<td>85 °C</td>
<td>400 °C/3min</td>
<td>RT</td>
</tr>
<tr>
<td>Q100P</td>
<td>1100 °C/15min</td>
<td>100 °C</td>
<td>400 °C/3min</td>
<td>RT</td>
</tr>
<tr>
<td>Q115P</td>
<td>1100 °C/15min</td>
<td>115 °C</td>
<td>400 °C/3min</td>
<td>RT</td>
</tr>
<tr>
<td>Q150P</td>
<td>1100 °C/15min</td>
<td>150 °C</td>
<td>400 °C/3min</td>
<td>RT</td>
</tr>
<tr>
<td>Q170P</td>
<td>1100 °C/15min</td>
<td>170 °C</td>
<td>400 °C/3min</td>
<td>RT</td>
</tr>
</tbody>
</table>
and at 130° (220) and 79° (200) with an exposure time of 30 s for austenite. The peaks were analysed using XTronic 1.12.0 -software, and the retained austenite fractions were calculated from integrated peak intensities as described in the standard [30]. Additionally, the α'-martensite volume fractions of the samples were measured with magnetic characterization device Feritscope® FMP30 before and after the cavitation tests. The Feritscope® was calibrated with three δ-ferrite standard samples, and a pre-set correction factor was used for calculating the α'-martensite fraction [31].

3. Results and discussion

3.1. Material properties after heat treatments

The retained austenite analysis with the chosen XRD-method is based on calculating the integrated intensities of two ferrite and two austenite peaks, and comparing these intensities to each other as described in standard [30]. Examples for the acquired peaks are shown in Fig. 2. The used peaks in Fig. 2 are located in the middle of each detector. The measured retained austenite fractions are shown in Fig. 3. The measured values were corrected for carbides by adding carbide fractions to the calculation as described in authors’ earlier work [17]. Although XRD-measurements are susceptible to sample preparation and presence of other phases that overlap with the chosen peaks, the measurements provide comparable data between different samples. Partially overlapping carbide peaks, however, can be seen to produce rather large error. Based on the XRD-data it can be observed that sample Q100P has the highest retained austenite fraction. After initial quenching to 100 °C the carbon partitioning is sufficient to stabilize the remaining austenite upon final quenching. As discussed in Ref. [17], altering the initial quenching temperature from the optimal range results in lower retained austenite fraction. Increasing the initial quenching temperature above 100 °C reduces the retained austenite fraction gradually, as less carbon is available for stabilizing the austenite phase upon final quenching. At lower quenching temperatures the martensitic transformation advances rapidly to a robust level, and the retained austenite fraction stays at roughly 15% independent on the following partitioning or tempering heat treatment.

As the merely quenched sample DQ also contains retained austenite, the Ms temperature of this steel grade is below room temperature. This was also confirmed with dilatometry analysis as shown in Fig. 4. The Ms temperature of the steel was determined to be 175 °C, though the martensitic transformation can be observed to begin slowly. Therefore, it is likely that carbon partitioning to austenite was incomplete after quenching to high temperatures, and the Q&P samples Q115P, Q160P, and Q170P contain fresh martensite that was formed upon the final quenching. However, when martensite is formed before the subsequent partition step, some stress relieving can occur as martensite loses its tetragonality during partition. This should be considered especially for samples Q85P and Q100P. Moreover, as both samples DQ and Q&T were initially quenched to room temperature, it is possible that the quenched sample DQ contains more retained austenite than was measured with XRD. The low retained austenite fraction measured could be caused by sample preparation, since the sample DQ is very metastable due to lack of subsequent partitioning or tempering heat treatment. As metastable austenite is prone to phase transformation [19,32–34], some retained austenite at the surface layer could transform to martensite during

Fig. 2. Examples of XRD-peaks for a) sample DQ, and b) sample Q100P. Ferrite peaks Ferr_1 and Ferr_2 correspond to 2θ-peak angles 156.4° (211) and 106.1° (200), and austenite peaks Aust_1 and Aust_2 angles 130° (220) and 79° (200), respectively.

Fig. 3. Retained austenite fractions (RA-%) in volume-% with error bars from XRD-measurements.

Fig. 4. Ms, As, and Af temperatures determined by dilatometry. The end temperature was set to room temperature, and Ms temperature was not reached.
The analysis of the initial microstructures is supported by the hardness measurements presented in Fig. 5. The highest hardness values were measured in samples DQ, Q115P, Q160P, and Q170P, where the hardness of the steel is roughly 660 HV2. This indicates the presence of fresh martensite in these samples. Although the steel has been reported to contain up to 5% carbides after Q&P [17], their effect on the total hardness is insignificant. The softest samples were Q&T and Q100P, although their lower hardness is caused by different reasons: The lowest measured hardness of sample Q&T shows that the martensite has been tempered during the 2-h-hold at 200 °C. The lower hardness of Q100P, however, can be explained by its higher retained austenite fraction, although the loss of martensite tetragonality during partition provides additional softening effect. Interestingly, samples Q50P200 and Q85P have very similar initial properties despite differing heat treatments. Although the sample Q85P has slightly higher retained austenite fraction, it is still slightly harder than Q50P200. This suggests that the martensite in Q50P200 is softer due to tempering during the longer partition time.

3.2. Mass loss measurements

Fig. 6 shows the cumulative mass losses of the samples as a function of exposure time as average of three measurements. In standard G32-16 [29], the preferred primary result for erosion test would be mean depth of erosion, which could be derived from the mass loss by dividing it with sample density and test surface area. However, as all samples of this study are of the same material and share the same density and test area, mass loss data will provide similar comparative results. As can be seen from Fig. 6, the initial mass loss is very high in all samples, after which the mass loss stabilizes to a steadier rate. The short initial acceleration stage in cavitation erosion can be explained by limited plastic deformation due to phase transformations which occurred during the sample preparation [35–37]. The mass loss rate is then reduced, as strain induced martensitic transformation can absorb some of the impact energy upon collapse of cavitation bubbles [20–24]. The weakest performance of the quenched and tempered sample is most likely caused by the softer martensitic microstructure after tempering. As sample Q50P200 was also partitioned for a long time (1 h) at the same temperature, and it performed only slightly better than Q&T-sample, the results suggest that tempering of martensite reduces the cavitation erosion resistance significantly. This is also supported by the remarkably lower mass loss of quenched sample DQ, which had no tempering or partitioning treatment before the cavitation test. Quenched and partitioned samples, however, showed clearly enhanced resistance to cavitation erosion. The best performing sample was Q100P, which had the highest initial retained austenite fraction.

3.3. Morphologies of eroded surfaces

The surface morphologies of eroded samples DQ, Q50P200, Q&T, Q100P, and Q170P are presented in Fig. 7. The least eroded sample Q100P contained areas with little to no wear (Fig. 7d) and areas with slight wear at the grain boundaries (Fig. 7e). As can be observed from all surfaces, the grain boundary regions seem to be the locations for initial cavitation erosion. Grain boundary fracture has been reported [22] as the main damage mechanism in other stainless steels as well, after which the erosion expands to the martensite grains surrounding the boundary region. In samples Q&T and Q50P200, with longer partition/tempering time at lower temperature, the cavitation erosion has proceeded inside the martensitic grains and heavy erosion can be observed throughout the surface. This is likely caused by the softer matrix consisting of tempered martensite. After 6 h some parts of the grain surface remain intact, however, and the surface structure is not completely eroded. The coarse structure of eroded areas suggests that the failure mechanism upon material removal is mostly brittle peeling of small pieces at the cracks. As the material has been reported [17] to contain carbides after Q&P, dark round pits visible on the surfaces are formed by removal of carbide particles. This is especially visible in sample Q170P in Fig. 7f.

In addition to electron microscopy, the surfaces were analysed using optical profilometry. These results for samples DQ, Q85P, Q100P, Q170P, Q50P200, and Q&T are shown in Fig. 8. As can be observed in Fig. 8, the middle of the eroded areas has expanded in all samples resulting in increased height compared to the polished area. An example of this is further highlighted by a cross-section analysis in Fig. 9. Purple color-range at the edges of the cavitated areas correspond to heavily eroded regions common to this test method [10,38]. In sample DQ the heavy cavitation erosion at the edge has led to formation of cracks that can be observed with optical profilometry in Fig. 8a.

3.4. Correlation between cavitation erosion and microstructure

Based on total mass losses in Fig. 6, the cavitation erosion resistance of the steel is mainly dependent on the strength of the martensitic matrix. Due to stress relieving and softening of the martensite during tempering or longer partition, the strength of the matrix is lowered resulting in lower resistance to cavitation erosion. Directly after quenching the matrix comprises mainly fresh martensite and some retained austenite. This kind of structure is very hard, which results in decent cavitation erosion resistance. However, it is also very brittle and has tendency of cracking under stress. Despite having better cavitation erosion resistance than tempered samples after 6 h, the cracks observed on the surface in Fig. 8a indicate that the steel might not perform well after longer test time if the crack propagation advances.

The Q&P-samples that were partitioned for 3 min at 400 °C performed the best in this study. By comparing the cumulative mass losses...
in Fig. 6 to initial hardness and retained austenite measurements in Figs. 3 and 5, the initial properties can be associated with the severity of cavitation erosion. This comparison is illustrated in Fig. 10. The most important factor affecting the cavitation erosion resistance is the hardness of the martensitic phase. This can be observed by comparing samples Q&T, Q50P200, Q100P, and Q170P in Fig. 10, for example. As sample Q&T was tempered for the longest time, its martensitic matrix is the softest of all samples, which explains the highest mass losses during
cavitation erosion. For sample Q50P200, the low-temperature partitioning was much shorter, and the martensitic matrix was therefore harder than in quenched and tempered sample. Although samples Q100P and Q170P differ significantly in measured hardness, that difference is caused by differences in their phase fractions. Sample Q100P contains the largest fraction of soft austenitic phase, which drastically reduces the measured hardness of the steel. However, due to similar partitioning heat treatment, the martensitic phase is very hard in both of these samples. This explains the relatively small difference in mass losses between Q&P samples partitioned at 400 °C in contrast to the tempered samples.

The initially softest sample Q100P performed significantly better than any other sample. The high performance of Q100P can be explained by its highest retained austenite fraction, which enables energy absorption through strain induced martensitic phase transformation. This phase transformation causes internal compressive stress within the lattice due to expansion of volume. Although earlier studies [14,17] show that the examined steel performs poorly under tensile load, it has been reported to perform well under compression [39]. The weak tensile properties were attributed to cracking at grain boundaries [17], which also serve as nucleation sites for cavitation erosion as observed in Fig. 7.

Hence, it is likely that the compressive stresses arising from the martensitic transformation suppress the propagation of grain boundary fractures. This behaviour has also been observed in an earlier wear related study by Saha et al. [40]. Having high retained austenite fraction therefore provides a source for internal compressive stress which delays the cavitation erosion at grain boundaries. Additionally, the formation of strain induced martensite increases the hardness of the grains which further reduces cavitation erosion.

The amount of transformed martensite was measured with Feritscope® as shown in Fig. 11. Although Feritscope® has its limitations due to calibration with ferrite samples and the use of correction factors for martensite [31,41], it shows clearly that the α'-martensite fraction was increased in almost all samples during the cavitation tests. This increase caused by strain induced martensitic transformation can be used for comparing samples after the heat treatments. The highest increase of α'-martensite fraction can be observed in samples Q&T, Q50P200, and Q100P. These samples contained the least α'-martensite before the cavitation tests and had high initial fractions of retained austenite. Samples Q&T and Q50P200 contain mostly tempered martensite which results in lower magnetic response than α'-martensite, whereas in sample Q100P the low initial α'-martensite fraction results from its high retained austenite fraction. During cavitation erosion some of the retained austenite is transformed to martensite, which results in increasing magnetic response. Hence, a paramagnetic phase is replaced by a ferromagnetic one. As Q100P initially contains α'-martensite likewise to other Q&P-samples, the observed increase in α'-martensite fraction is slightly smaller than in samples containing tempered martensite.

When α'-martensite fraction approaches 85 vol.-%, there is little difference between the measurements before and after cavitation test. This is partially due to the detection limit of the Feritscope® at high α'-martensite fractions, but it is also dependent on the initial state of the material after sample preparation. It is possible that, despite different measured retained austenite fractions with XRD, samples DQ, Q115P, Q160P, and Q170P actually contain approximately 15 vol.-% of retained austenite. As these samples were eventually quenched to room temperature, based on the dilatometry curve in Fig. 4 it would be unlikely that they would contain so little retained austenite. Considering the surface had likely undergone some martensitic transformation during the sample preparation, the initially measured values may differ from what would be measured from the inside of the sample. Both hardness measurements and Feritscope® measure the material properties from deeper depth than XRD, which would explain the smaller differences observed with these methods: the resulting graphs between Figs. 5 and 11 are very similar. Thus, Q&P samples initially quenched to 115 °C or higher would produce similar microstructures. Their better performance compared to sample DQ can be explained by their short partitioning at 400 °C where some of the tetragonality of the martensite is lost, resulting
in increased ductility.

Another factor influencing the results between the samples could be austenite stability due to different heat treatment parameters used. By applying the constrained carbon equilibrium model of Speer et al. [12] as described in authors’ earlier work [17], the carbon content of the retained austenite can be estimated for different samples. The lower the initial quenching temperature is, the more carbon is available for stabilizing the retained austenite during subsequent partitioning or tempering heat treatment. As a result, the retained austenite in samples DQ, Q115P, Q160P, and Q170P would have very low stability, and fresh martensite is easily formed during deformation. Contrarily, in samples Q&T, Q50P200, and Q85P the retained austenite carbon content should be the highest, resulting in higher austenite stability. Thus, more energy would be needed in these samples for martensitic transformation, resulting in weaker TRIP-effect. This could explain the poorer performance of sample Q8SP in comparison with other samples partitioned at 400°C. Based on the Feritscope® measurements in Fig. 11 and profilometry data in Fig. 8, however, the overall retained austenite stability of studied samples seems rather low, and fresh martensite is formed within the samples during cavitation erosion.

During cavitation erosion, the transformed surface layer is partially removed, and fresh strain induced martensite forms at the surface by a TRIP-mechanism. Hence, despite some retained austenite will be transformed to martensite during cavitation erosion, it will mostly replace the previously eroded layer, and little difference in measured α′-martensite fraction can be observed. This indicates that the phase transformation actively replaces the eroded layer with strain induced martensite during cavitation. This increases the hardness of the microstructure, which effectively reduces the erosion of the surface. Moreover, as the cavitation erosion damage reaches only few micrometers from the surface, as observed in Fig. 9, the material deeper under the surface is likely to remain unaffected and can provide hardening through a TRIP-mechanism over time as the damage layer proceeds further into the material.

4. Conclusions

This work examined the enhancement of cavitation erosion resistance of AISI 420-type stainless steel with Q&P. The effect of Q&P parameters and comparison of properties to quenched and tempered sample were discussed. Based on the results, the following conclusions were drawn.

1. The cavitation erosion resistance of EN 1.4034 stainless steel can be significantly enhanced with quenching and partitioning. The final properties, however, are highly dependent on the used heat treatment parameters. Low temperature tempering or partitioning results in soft tempered martensite matrix, whereas short partitioning at high temperature results in formation of hard α′-martensite upon final quenching.

2. The resistance to cavitation erosion is dependent on combination of the hardness and the retained austenite fraction of the steel. Increasing the hardness of the martensitic matrix generally improves the cavitation erosion resistance, whereas softer retained austenite provides TRIP-mechanism which further reduces cavitation erosion.

3. High retained austenite fraction is favourable, as strain induced martensitic transformation absorbs the cavitation damage upon plastic deformation. Lattice expansion caused by the phase transformation amplifies the compressive stresses generated by cavitation, which hinders the propagation of cracks at crack boundaries. Additionally, the resulting increase in α′-martensite fraction increases the hardness of the steel which further improves the cavitation erosion resistance.

CRediT authorship contribution statement

Lassi Raami: Conceptualization, Methodology, Investigation, Writing - original draft, Funding acquisition. Tommi Varios: Conceptualization, Validation, Writing - review & editing. Kati Valtonen: Investigation, Validation, Writing - review & editing. Marco Wendler: Validation, Writing - review & editing, Project administration. Olena Volkova: Validation, Writing - review & editing, Project administration. Pasi Peura: Validation, Resources, Writing - review & editing, Supervision, Project administration, Funding acquisition.

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Declaration of competing interest

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

Data availability

The raw/processed data required to reproduce these findings cannot be shared at this time as the data also forms part of an ongoing study.

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