Effects of Flow Velocity and Sand Concentration on Hydro-Abrasive Erosion Behavior of HVOF-Sprayed WC-Cr₃C₂-Ni Coating in 3.5 wt.% NaCl Aqueous Solution

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Abstract: To explore the hydro-abrasive erosion (HAE) behaviors of a high-velocity oxygen fuel (HVOF)-sprayed WC-Cr₃C₂-Ni coating in 3.5 wt.% NaCl aqueous solution, various flow velocities (FVs) and sand concentrations (SCs) were designed using a rotating disk rig facility. In comparison with 1Cr18Ni9Ti stainless steel, the WC-Cr₃C₂-Ni coating possessed higher hardness (H) and elastic modulus (E) values, worse anticorrosion properties, and superior HAE resistances in 3.5 wt.% NaCl aqueous solution. Higher FVs and SCs caused more severe HAE degradations for both the WC-Cr₃C₂-Ni coating and the 1Cr18Ni9Ti stainless steel, while FV had a greater influence on HAE resistances than SC. With an increase in the FV and SC, uncontinuous corrosion product films and erosion pits, the micro-cutting of the soft binder matrix and the fracturing of hard-phase grains, and crater formation and coating spalling were addressed as the HAE failure mechanisms in the 3.5 wt.% NaCl aqueous solution of the WC-Cr₃C₂-Ni coating.

Keywords: hydro-abrasive erosion; flow velocity; sand concentration; HVOF; WC-Cr₃C₂-Ni coating

1. Introduction

Improving the hydro-abrasive erosion (HAE) resistance of existing seawater pump impellers has been a matter of the ongoing effort to improve nuclear power plants. Except for the selection and the design of the pumps and the optimization of operating conditions, different types of surface coatings, such as iron-based amorphous coatings [1,2], nickel-based coatings [3], cobalt-based high-entropy coatings [4], and carbide-based cermet coatings [5–7], have shown the potential of protecting flow passage components. Among these, thermal-sprayed carbide-based cermet coatings possess superior hardness, adequate toughness, and excellent chemical stability, and, thus, they have been widely employed as reinforcement for seawater pump impellers [8–10].

However, there may exist the production of brittle phases, a decrease in adhesive strength, and the degradation of cohesion for carbide-based cermet coatings, owing to the decarburization and oxidation of carbide hard phases in the process of thermal spraying [11,12]. At present, the high-velocity oxygen fuel (HVOF) spraying technique predominates in the preparation of carbide-based cermet coatings. This technique promises low porosity, few brittle phases, and high bond strength for carbide-based cermet coatings compared to other thermal spraying approaches [13,14]. Thus, HVOF-sprayed carbide-based cermet coatings can provide promising erosion wear resistance after long-term exposure to seawater slurry environments.

There have been many studies on HVOF-sprayed WC-Cr₃C₂-Ni coatings, mainly regarding their wear and erosion performances, which are highly promising for the feasibility of protection against HAE. Compared with HVOF-sprayed WC-CoCr coatings, a lower friction coefficient and higher sliding wear resistance at various load or sliding
speed conditions have been measured for HVOF-sprayed WC-Cr$_3$C$_2$-Ni coatings as a consequence of their double-decker shell–core microstructures and well-distributed thin oxide scales, especially when the temperature exceeded 700 °C [15,16]. Interestingly, some studies have reported that the effect of the content of the alloying element Cr on the sliding wear resistance of the coating is not distinct when sliding against the WC-Co hard metal counterpart [17]. Moreover, it has been reported that low porosity, as well as high hardness and bond strength, could be achieved by optimizing the HVOF spraying process and the feedstock powder characteristic, as well as by the heat treatment of the HVOF-sprayed WC-Cr$_3$C$_2$-Ni coating, causing enhanced sliding wear and abrasive wear resistances [18–21]. With regard to erosion, both the environmental conditions and flow velocity (FV) have been found to significantly influence the erosion performance of HVOF-sprayed WC-Cr$_3$C$_2$-Ni coatings [22–24]. Different from cavitation erosion [25–27], HAE behavior is not only correlated to the impact angle and velocity of the silt-laden water flow, but it is also affected by the concentration and grainsize of sand particles. To the best of our knowledge, there has been no careful evaluation of the HAE behavior of HVOF-sprayed WC-Cr$_3$C$_2$-Ni coatings in seawater slurries.

In this work, a WC-Cr$_3$C$_2$-Ni cermet coating was manufactured using the HVOF spraying technique, and its HAE properties in 3.5 wt.% NaCl aqueous solution were analyzed by taking advantage of a rotating disk rig facility. Meanwhile, the influences of the FV and sand concentration (SC) on the HAE behavior and failure mechanism were discussed on the basis of volume loss rates and eroded surface morphologies.

2. Materials and Methods

The feedstock powder and substrate material used for spraying in the present research were commercially available WC-Cr$_3$C$_2$-Ni powder obtained from Chongyi Zhangyuan Tungsten Co., Ltd. (Chongyi, China) and 1Cr18Ni9Ti stainless steel, respectively. Their chemical compositions (wt.%) were 68.53 W, 17.33 Cr, 7.14 C, 7.0 Ni, and 69.63 Fe, 17.65 Cr, 10.23 Ni, 1.17 Mn, 0.75 Si, 0.46 Ti, 0.075 C, 0.021 P, 0.0069 S, respectively. 1Cr18Ni9Ti stainless steel was also selected as the comparative material for the HAE tests. WC-Cr$_3$C$_2$-Ni feedstock powders were sprayed onto the substrates with the aid of Tafa-JP8000 HVOF spraying equipment manufactured by Praxair Co., Ltd. (Danbury, CT, USA) after the preliminary treatment of the substrates, which included ultrasonic cleaning, drying, and sandblasting. The specific HVOF spraying parameters were as follows: a kerosene flow rate of 23.4 L·h$^{-1}$, an oxygen flow rate of 906 L·min$^{-1}$, a spray distance of 0.38 m, a powder feeding rate of 0.92 g·s$^{-1}$, and a spray gun scanning speed of 16.8 m·min$^{-1}$. Ar was introduced as the powder carrier gas with a flow rate of 652 L·h$^{-1}$. The coatings were ~350 µm thick, and the number of spraying passes was equal to 14.

The X-ray diffraction method (XRD, Bruker D8-Advanced, Bruker, Karlsruhe, Germany) with a Cu Kα anode was utilized to analyze the phase composition of the WC-Cr$_3$C$_2$-Ni feedstock powders and as-sprayed coatings. Field-emission scanning electron microscopy (FE-SEM, Zeiss Gemini Sigma 300, Zeiss, Jena, Germany) was employed to observe the surface and cross-sectional morphologies of the WC-Cr$_3$C$_2$-Ni feedstock powders and as-sprayed coatings. Moreover, an element analysis of the eroded surfaces was performed using energy-dispersive spectroscopy (EDS, Bruker XFlash 6160, Bruker, Karlsruhe, Germany). Porosity testing was conducted under a magnification of 1000 on the FE-SEM cross-sectional morphologies of the as-sprayed coatings via the image analysis method. The nanoindentation method on the cross-section of the as-sprayed coating and the comparative stainless steel via an ultra nanoindentation tester (CSM UNHT, Peseux, Switzerland) with a Berkovich tip was used to assess the mechanical properties, comprising the hardness ($H$) and elastic modulus ($E$). During mechanical data acquisition, the maximum load, duration time of the maximum load, strain rate, loading rate, and unloading rate were set as 300 mN, 5 s, 0.05 s$^{-1}$, 10 mN·s$^{-1}$, and 10 mN·s$^{-1}$, respectively. The porosity and nanoindentation tests were both completed for 15 repetitive measurements to ensure the accuracy of the data. The electrochemical corrosion of the as-sprayed coating and the comparative stainless steel
were tested using a CS 2350H electrochemical workstation (Corrtest Instrument Co., Ltd., Wuhan, China). Potentiodynamic polarization tests were performed using a typical three-electrode electrochemical cell, in which a platinum plate, a saturated calomel electrode (SCE), and a test specimen served as the auxiliary electrode, the reference electrode, and the working electrode, respectively. Each test specimen was put into 3.5 wt.% NaCl aqueous solution at room temperature for 3600 s to reach the equilibrium state of the open-circuit potential. After that, potentiodynamic polarization curves were recorded at a scanning speed of 0.5 mV·s⁻¹, and the resultant electrochemical characteristic parameters, including corrosion current density ($i_{corr}$) and corrosion potential ($E_{corr}$), were obtained.

HAE tests were conducted using a rotating disk rig facility with a circulating system, as illustrated in Figure 1a. The schematic illustrations of the rotating disk compartment for the HAE tests and the corresponding specimen size are shown in Figure 1b,c, respectively. Six radial holes along the circumference of the rotating disk were designed for the test specimens. The high-velocity sand-containing seawater in the Qinshan area of China contains an average flow velocity of 0.9–1.8 m·s⁻¹ with a maximum of 4 m·s⁻¹ and an average sand concentration of 3–6 kg·m⁻³ with a maximum of 12 kg·m⁻³ [1,28]. In order to forecast the performance of the HVOF-sprayed WC-Cr₃C₂-Ni coating for application in seawater pumps, the operating parameters listed in Table 1 and 3.5 wt.% NaCl aqueous solution were selected to simulate the severe working environments that run under highly HAE conditions in corrosive seawater. The preliminary treatment of the test specimens, including grinding using 240–2000# SiC abrasive papers, polishing using 1.5 µm diamond pastes, cleaning ultrasonically in acetone, drying in hot air, and weighing using an analytical balance with a precision of 10⁻⁴ g, was conducted. After that, the test specimens were embedded in the rotating disk, and the HAE tests were performed for 6 h in a slurry solution containing 3.5 wt.% NaCl and commercial quartz sand, with a particle size range of 30–70 mesh. A cooling circulating system was employed to maintain the temperature of the slurry solution at 25 ± 5 °C. To ensure the accuracy of the data, new quartz sand and three parallel specimens for the coating and the comparative stainless steel were used at each operating parameter. After the end of each test, the specimens were re-treated following the procedure used to determine the mass changes, which included cleaning ultrasonically in acetone, drying in hot air, and weighing using an analytical balance. The volume losses were calculated by dividing the mass losses by the densities of the specimens. The densities of the WC-Cr₃C₂-Ni coating and the 1Cr18Ni9Ti stainless steel were determined using the buoyancy method, and they were 11.83 and 7.93 g·cm⁻³, respectively. With regard to the coating, the density was determined using buoyancy measurements on free-standing coatings. Thus, the volume loss rates and the corresponding FE-SEM morphologies of the eroded surfaces were mainly considered to reveal the effects of FV and SC on the HAE behaviors in the 3.5 wt.% NaCl aqueous solution of the HVOF-sprayed WC-Cr₃C₂-Ni coating. The surface roughness ($R_a$) of the eroded surfaces was also measured using a laser scanning confocal microscope (LSCM, Olympus OLS4100, Olympus, Tokyo, Japan).

Table 1. Operating parameters of the HAE tests in 3.5 wt.% NaCl aqueous solution.

| Operating Parameters | Flow Velocity (m·s⁻¹) |
|----------------------|-----------------------|
|                      | 23.4                  | 33.5                  | 41.9                  |
| Sand concentration (kg·m⁻³) | 10                    | √                     | √                     |
|                      | 20                    | √                     | √                     |
|                      | 35                    | √                     |                       |
which was unfavorable for the HAE resistance of the coating to a certain degree.

which guaranteed good fluidity and sufficient combustion of the feedstock powders for

µ were spherical with a uniform size (15~45 µm) were present in the feedstock powder and the coating. It can be observed in Figure 2a that WC appeared as the major phase, and Cr$_3$C$_2$ and Ni were the minor phases in the feedstock powder. However, the intensity of the peaks corresponding to WC, Cr$_3$C$_2$, and Ni became weaker, and the peaks corresponding to (W,Cr)$_2$C appeared after the HVOF spraying process. These findings indicate that Cr$_3$C$_2$ was unstable along with the Ni matrix in the molten state, especially when WC, Cr$_3$C$_2$, and Ni sufficiently underwent solid-state reactions due to the sufficiently high temperature and complete combustion [29–31]. The powder manufacturing method contained the process of ball-milling in deionized water and the open spray-drying system, which helped to maintain a low oxygen content in the agglomerated powders. As can be seen in Figure 2b, the spray-dried feedstock powders were spherical with a uniform size (15~45 µm) and rarely observable pores on the surface, which guaranteed good fluidity and sufficient combustion of the feedstock powders for the duration of the HVOF spraying process [32]. For the coating in the as-sprayed state as shown in Figure 2c–e, the structure was rather dense, and few micro-pores were found among disparate hard phases and binder phases. Apart from this, the thickness and the average porosity of the coating were 350 µm and 1.1%, respectively. It can also be noted that there was good bonding between the coating and the substrate, as well as between the tungsten carbides and chromium carbides. These observations suggest that sufficient deformation and the anchoring effect took place due to the high-speed flight and severe impact of the feedstock powders in the molten state [24].

Table 2 lists the mechanical properties and some electrochemical characteristics of the WC-Cr$_3$C$_2$-Ni coating and the 1Cr18Ni9Ti stainless steel. It can be noted that the $H$ and $E$ values of the coating were 3 times and nearly 2 times higher than those of the stainless steel, respectively, which can facilitate the enhancement of the HAE resistance of the coating. By contrast with the stainless steel, the superior mechanical properties of the coating were not only caused by the high proportion of tungsten carbide and chromium carbide phases, but they were also related to the few internal defects in the coating [33,34]. It can also be noted that the $E_{corr}$ value of the coating appeared to be more negative by comparison with the stainless steel, and the $i_{corr}$ value of the coating was about 189 times larger than those of the stainless steel in the 3.5 wt.% NaCl aqueous solution. These findings indicate that the coating possessed worse anticorrosion properties in the 3.5 wt.% NaCl aqueous solution, which was unfavorable for the HAE resistance of the coating to a certain degree.
Figure 2. Phase constitutions and microstructures of the feedstock powder and the coating: (a) XRD patterns [23]; (b) surface morphology of the feedstock powder; (c, d) cross-sectional morphology of the coating; (e) porosity of the coating.

Table 2. Mechanical properties and some electrochemical characteristics.

| Materials             | WC-Cr$_3$C$_2$-Ni Coating | 1Cr18Ni9Ti Steel |
|-----------------------|---------------------------|------------------|
| Hardness ($H$), GPa   | 11.9 ± 1.7                | 2.9 ± 0.4        |
| Elastic modulus ($E$), GPa | 323 ± 29                   | 117 ± 14         |
| Corrosion potential ($E_{corr}$), mV vs. SCE | −510                     | −204             |
| Current density ($i_{corr}$), nA·cm$^{-2}$ | 7480                     | 39.5             |

Figure 3 shows the VLR of the WC-Cr$_3$C$_2$-Ni coating and the 1Cr18Ni9Ti stainless steel as a function of the FV after HAE at 20 kg·m$^{-3}$ in the 3.5 wt.% NaCl aqueous solution. After 6 h of HAE at FVs of 23.4, 33.5, and 41.9 m·s$^{-1}$, the coating revealed VLR values of 0.005, 0.016, and 0.025 mm$^3$·h$^{-1}$, respectively, which were dramatically lower than those of the stainless steel (1.0, 3.5, and 5.8 mm$^3$·h$^{-1}$). In addition to this, the VLR values of the coating and the stainless steel after 6 h of HAE at the FV of 41.9 m·s$^{-1}$ were 4 times and nearly 5 times higher than those after 6 h of HAE at the FV of 23.4 m·s$^{-1}$, respectively. It should be mentioned that more severe HAE degradation in the 3.5 wt.% NaCl aqueous solution took place for both the coating and the stainless steel at a higher FV coupled with the SC of 20 kg·m$^{-3}$. When the FV increased from 23.4 m·s$^{-1}$ to 41.9 m·s$^{-1}$, there was a larger surface area for electrochemical reactions, a wider availability of sand particles, and higher stress on the surface of the specimen as a consequence of the faster conveyance of electroactive species, as well as the higher flow turbulence and kinetic energy conveyed by sand particles [35-37]. Meanwhile, the frequency of sand particle impingement increased as the FV increased, leading to a weaker shielding effect [38]. These reasons account for the more severe HAE degradation in the 3.5 wt.% NaCl aqueous solution.
After 6 h of HAE in the 3.5 wt.% NaCl aqueous solution at 20 kg·m⁻³ and different FVs, the FE-SEM images of the surfaces of the WC-Cr₃C₂-Ni coatings are shown in Figure 4. After HAE at FVs of 23.4, 33.5, and 41.9 m·s⁻¹, the R₄ values of the coatings were 0.14, 0.23, and 0.29 μm, respectively. With an increase in the FV, the eroded surfaces gradually became more irregular and rougher since there were larger numbers and dimensions of erosion pits or craters, more obvious scour marks, and a smaller area of initially polished surfaces, as revealed in Figure 4a–c. During HAE at the higher FV, more serious degradation of the coating took place because of the occurrence of more drastic physicochemical interaction processes, which included the impingement and abrasion of sand particles, as well as electrochemical reactions. This contributed to the increase in the VLR value as stated in Figure 3. At the FV of 23.4 m·s⁻¹ coupled with the SC of 20 kg·m⁻³, the scattered distribution of small erosion pits, combined with the slight fracturing of hard-phase grains and the micro-cutting of the soft binder matrix, was detected on the relatively flat surface, as shown in Figure 4a₁,a₂. These observations indicate that a large proportion of the interfaces between the hard-phase grains and the soft binder matrix were sufficient to resist the impingement and abrasion of sand particles at relatively low FVs. From Figure 4b₁,b₂, it was confirmed that there were shallow erosion pits with larger diameters of approximately 5 μm when the FV ranged from 23.4 m·s⁻¹ to 33.5 m·s⁻¹, although most of the hard-phase grains and the soft binder matrix were still relatively intact. With a further increase in the FV, both the dimension and the depth of the craters were larger, as presented in Figure 4c₁,c₂. Additionally, a somewhat interconnected groove structure was formed due to the gradual emergence of crater-shaped structures when the FV was 41.9 m·s⁻¹. Not only is this correlated with the more violent flow structure, but it is also because of the higher propagation rate of cracks at the higher FV of 41.9 m·s⁻¹ [39].

In Figure 5, the VLR of the WC-Cr₃C₂-Ni coating and the 1Cr18Ni9Ti stainless steel with respect to the SC after HAE at 33.5 m·s⁻¹ in the 3.5 wt.% NaCl aqueous solution is provided. After 6 h of HAE, the difference was obvious when the SCs were 10, 20, and 35 kg·m⁻³, in which the VLR values of the coating and the stainless steel were 0.008, 0.016, and 0.031 mm³·h⁻¹, and 3.1, 3.4, and 3.6 mm³·h⁻¹, respectively. Apart from this, the VLR values of the coating and the stainless steel after 6 h of HAE at the SC of 35 kg·m⁻³ were approximately 287% and 16% higher than those after 6 h of HAE at the SC of 10 kg·m⁻³, respectively. It was confirmed that more severe HAE degradation in the 3.5 wt.% NaCl aqueous solution occurred for both the coating and the stainless steel at higher SCs coupled with the FV of 33.5 m·s⁻¹ due to larger interactions between the sand particles and the specimen surface, resulting from more sand particles impinging on the specimen surface, as well as shorter intervals among different sand particle impingements [36,40].
After HAE at SCs of 10, 20, and 35 kg·m⁻³ (Figure 6a, a1), it was found that the microstructure of the coating was fairly compact, exhibiting the characteristics of the slight micro-cutting of the soft binder matrix and the fracturing of hard-phase grains, as well as a few scattered erosion pits with a diameter of around 1.2 μm, which provided evidence for the strong interface bonding between the hard phase and the binder matrix ensured the sufficient propagation of cracks. Moreover, at higher FVs and SCs, the coating's eroded surface exhibited more obvious fracturing of hard-phase grains, micro-cutting, deeper erosion pits, and a few scattered erosion pits, which mean that corrosion product films were fractured and the formation of debris structures occurred due to the gradual emergence of prolonged cracks in the coating.

Comparisons among the EDS results of Points A, B, and C can reflect the destruction of the corrosion product films with a higher probability at higher FVs and SCs, since the content of the O element in the bright white structure after being subjected to higher FVs and SCs was substantially increased (Figure 4b, b1, b2). This increase is evidenced by the measured VLR values, as mentioned in Figure 5. Notably, the existence of oxides existed in bright white structures, which means that corrosion product films were approximately 287% and 16% higher than those after 6 h of HAE at the SC of 10 kg·m⁻³, respectively. It was confirmed that more severe HAE degradation in the 3.5 wt.% NaCl aqueous solution occurred for both the coating and the stainless steel at higher SCs and different FVs, as shown in Table 3. The R₄ values of the coatings were 0.17, 0.23, and 0.27 μm, respectively. Apart from this, the VLR values of the coating and the stainless steel were 0.008, 1.016, and 0.031 mm³·h⁻¹, respectively. According to the coating’s eroded surface at the SC of 10 kg·m⁻³ (Figure 6a, a1), it was found that the microstructure of the coating was fairly compact, exhibiting the characteristics of the slight micro-cutting of the soft binder matrix and the

Figure 6 illustrates the FE-SEM images of the surfaces of the WC-Cr₃C₂-Ni coating after 6 h of HAE in the 3.5 wt.% NaCl aqueous solution at 33.5 m·s⁻¹ and different SCs. After HAE at SCs of 10, 20, and 35 kg·m⁻³, the R₄ values of the coatings were 0.17, 0.23, and 0.27 μm, respectively. According to the coating’s eroded surface at the SC of 10 kg·m⁻³ (Figure 6a, a1), it was found that the microstructure of the coating was fairly compact, exhibiting the characteristics of the slight micro-cutting of the soft binder matrix and the
fracturing of hard-phase grains, as well as a few scattered erosion pits with a diameter of around 1 µm, which provided evidence for the strong interface bonding among the tungsten carbide, chromium carbide, and nickel binder. Specifically, the strong interface bonding between the hard phase and the binder matrix ensured the sufficient transfer of shear stress and normal stress at relatively low SCs, which suppressed the initiation of micro-cracks and promoted the increase in the HAE resistance of the coating. After being subjected to 20 and 35 kg·m⁻³ (Figure 6b,b1,c,c1), the compact microstructures of the coatings were not maintained, and they were replaced by larger and deeper erosion pits, the more obvious fracturing of hard-phase grains, micro-cutting, and the preferential removal of the soft binder matrix, as well as coating spalling. These observations indicate that higher SCs during HAE in the 3.5 wt.% NaCl aqueous solution intensified the degradation of the coating to a certain degree, which was also evidenced by the measured VLR values, as mentioned in Figure 5.

Figure 6. FE-SEM images of the WC-Cr₃C₂-Ni coatings after HAE for 6 h in 3.5 wt.% NaCl aqueous solution at 33.5 m·s⁻¹ and different SCs: (a) 10 kg·m⁻³; (a1) a magnification of the rectangular frame in (a); (b) 20 kg·m⁻³; (b1) a magnification of the rectangular frame in (b); (c) 35 kg·m⁻³; (c1) a magnification of the rectangular frame in (c).
EDS was used to further identify and compare the compositional change in the bright white phase, which was noticeable in all microstructures of the WC-Cr$_3$C$_2$-Ni coatings (Figures 4 and 6) after 6 h of HAE in the 3.5 wt.% NaCl aqueous solution at different FVs and SCs. Table 3 lists the analytical results of EDS. Notably, the existence of the O element was detected in all the microstructures of the coatings. It can be inferred that oxides existed in bright white structures, which means that corrosion product films were scattered on the eroded surfaces of the coatings. Comparisons among the EDS results of Points A, B, and C can reflect the destruction of the corrosion product films with a higher probability at higher FVs and SCs, since the content of the O element in the bright white structure after being subjected to higher FVs and SCs was substantially lower. Hence, with an increase in the FV and SC, the HAE process of the WC-Cr$_3$C$_2$-Ni coating in the 3.5 wt.% NaCl aqueous solution mainly included uncontinuous corrosion product films, erosion pits, the micro-cutting of the soft binder matrix and the fracturing of hard-phase grains, and crater formation and coating spalling.

| Region | O (at.%) | Cr (at.%) | C (at.%) | W (at.%) | Ni (at.%) |
|--------|---------|-----------|---------|---------|----------|
| A      | 9.6     | 14.6      | 57.8    | 12.5    | 5.5      |
| B      | 16.7    | 6.9       | 60.2    | 10.7    | 5.4      |
| C      | 8.4     | 3.3       | 62.7    | 22.7    | 3.0      |

4. Discussion

4.1. Influences of Microstructure, and Mechanical and Corrosion Properties on HAE

By comparing the VLR values of the WC-Cr$_3$C$_2$-Ni coating and the 1Cr18Ni9Ti stainless steel after HAE at different FVs and SCs as shown in Figures 3 and 5, the result that the coating possessed higher HAE resistance than the stainless steel in the 3.5 wt.% NaCl aqueous solution can be obtained. The first reason is that the dense structure of the coating and the strong interface bonding between the coating and the substrate, as well as among the tungsten carbide, chromium carbide, and nickel binder, can reduce the cracking sources and large-area exfoliation of the coating under the attack of multi-phase flow. The second reason is that the superior mechanical properties (e.g., H and E) of the coating, owing to the plentiful supply of tungsten carbide and chromium carbide hard phases, can guarantee a high resistance to the impingement and abrasion of sand particles, especially under normal impact [41–43]. The third reason is that the element chromium in the coating can promote the formation of passive films and mitigate the HAE degradation of the coating in the 3.5 wt.% NaCl aqueous solution to a certain degree [44,45], although the anticorrosion property of the coating was inferior to that of the stainless steel in the 3.5 wt.% NaCl aqueous solution. However, it should be noted that the 6 h-long experiment used in the present study may be too short to sufficiently capture the potentially important role of corrosion damage in the HAE response since there was a significant difference in corrosion susceptibility between the coating and the stainless steel. The intrinsic limitations of thermal-sprayed coatings, including the presence of micro-pores, as well as a relatively high surface roughness and weak inter-splat bonding, may result in increased corrosion susceptibility, the easier initiation of micro-cracks, and more serious HAE degradations, especially after long periods of exposure to a corrosive medium [46,47]. As actual service environments are less severe than those used in the present study, it is necessary to evaluate corrosion susceptibility in order to qualify the coatings for seawater pump applications.

4.2. Influences of FV and SC on HAE

By comparing the amplification of the VLR value of the WC-Cr$_3$C$_2$-Ni coating and the 1Cr18Ni9Ti stainless steel after HAE at different FVs and SCs as shown in Figures 3 and 5,
the result that the FV had a greater influence than the SC on HAE resistances for both the coating and the stainless steel can be obtained. This phenomenon happened mainly because of two reasons. On the one hand, SC only influenced the frequency of sand particle impingement, whereas FV could affect the kinetic energy conveyed by sand particles, the availability and shielding effect of sand particles, the conveying speed of electroactive species, and the surface area for electrochemical reactions, except for the frequency of sand particle impingement [35–38]. On the other hand, the increase in SC may result in a lower kinetic energy conveyed by sand particles due to the collision of different sand particles, especially when SC exceeded the critical value [48]. It is noteworthy that a higher FV coupled with an appropriate amount of SC may induce more serious HAE degradations.

5. Conclusions

WC-Cr$_3$C$_2$-Ni coatings were fabricated by HVOF spraying for resistance against HAE in 3.5 wt.% NaCl aqueous solution. The WC-Cr$_3$C$_2$-Ni coating presented a dense structure with a thickness of ~350 µm and an average porosity of 1.1%, as well as a good combination with the substrate. Compared to the 1Cr18Ni9Ti stainless steel, the WC-Cr$_3$C$_2$-Ni coating possessed higher H and E values, worse anticorrosion properties, and superior HAE resistances in 3.5 wt.% NaCl aqueous solution. Higher FVs and SCs caused more severe HAE degradations for both the WC-Cr$_3$C$_2$-Ni coating and the 1Cr18Ni9Ti stainless steel, while FV had a greater influence on HAE resistances for both the coating and the stainless steel than SC. With an increase in the FV and SC, the WC-Cr$_3$C$_2$-Ni coating experienced the HAE process in the 3.5 wt.% NaCl aqueous solution of uncontinuous corrosion product films and erosion pits; the micro-cutting of the soft binder matrix and the fracturing of hard-phase grains; and crater formation and coating spalling.

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Nomenclature

| Abbreviation | Description |
|--------------|-------------|
| HAE          | hydro-abrasive erosion |
| FV           | flow velocity |
| SC           | sand concentration |
| VLR          | volume loss rate |
| HVOF         | high-velocity oxygen fuel |
| H            | hardness |
| E            | elastic modulus |
| XRD          | X-ray diffractometer |
| FE-SEM       | field-emission scanning electron microscopy |
| EDS          | energy-dispersive spectroscopy |
| JCPDS        | Joint committee powder diffraction standards |
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