

Review

Wear Degradation of High-Nitrogen Austenitic Steels for Internal Combustion Engines in Hydrogen-Containing Alternative Fuels

Alexander I. Balitskii ^{1,2,*}, Valerii O. Kolesnikov ^{1,3}, Jarosław Chmiel ⁴, Valentina O. Balitska ⁵,
Marcin A. Królikowski ² and Ljubomyr M. Ivaskevych ¹

¹ Department of Strength of the Materials and Structures in Hydrogen-Containing Environments, Karpenko Physico-Mechanical Institute, National Academy of Sciences of Ukraine, 5 Naukova Str., 79601 Lviv, Ukraine; kolesnikov197612@gmail.com (V.O.K.); ivaskevych@ipm.lviv.ua (L.M.I.)

² Department of Mechanical Engineering and Mechatronics, West Pomeranian University of Technology in Szczecin, 19 Piastów Av., 70-310 Szczecin, Poland; marcin.krolkowski@zut.edu.pl

³ Faculty of Technologies and Information Systems, Luhansk Taras Shevchenko National University, 2 Viktor Novikov St., 37500 Lubny, Ukraine

⁴ Faculty of Economics and Transport Engineering, Maritime University of Szczecin, ul. Wały Chrobrego 1-2, 70-500 Szczecin, Poland; j.chmiel@pm.szczecin.pl

⁵ Department of Physics and Chemistry of Combustion, Lviv State University of Life Safety, 35 Kleparivska, 79000 Lviv, Ukraine

* Correspondence: abalicki@zut.edu.pl; Tel.: +48-91-449-4947

Abstract

The development of hydrogen and hydrogen-containing fuel infrastructure requires structural materials resistant to the combined effects of hydrogen exposure, cyclic contact stresses, and tribological loading. This study investigates hydrogen-assisted degradation of two cold-worked high-nitrogen Cr–Mn–N austenitic steels, DDT68 and P900, under dry rolling–sliding contact after controlled electrochemical hydrogen charging. Hydrogen uptake, wear intensity, microstructure, subsurface damage, and rolling contact fatigue (RCF) were analyzed comparatively. The mean grain size was approximately 25 μm for DDT68 and 43 μm for P900, while hydrogen concentrations were 15–17 and 12–14 ppm, respectively. Hydrogen increased the wear intensity of both steels, with wear acceleration factors ranging from approximately 2.1 to 10.0 for DDT68 and 1.4 to 4.2 for P900 over 200–600 N. The differences in wear intensity between DDT68 and P900 after hydrogen charging were statistically significant over the investigated load range. The finer-grained DDT68 exhibited greater hydrogen retention and a stronger hydrogen-associated wear response. The observed damage is consistent with HELP-assisted deformation localization, with possible contributions from HEDE and hydrogen–defect interactions, promoting subsurface microcracking, delamination, and spalling. Hertzian analysis located the maximum subsurface shear stresses approximately 0.035–0.062 mm below the contact surface. Finally, an indicator-based framework combining wear response, hydrogen content, and prescribed mass-loss reference levels is proposed for comparative durability assessment under the investigated conditions, rather than as a universally calibrated service-life prediction model.



Academic Editors: Jiaqiang E and Paulo Santos

Received: 10 July 2026

Revised: 8 September 2026

Accepted: 23 September 2026

Published: 4 October 2026

Copyright: © 2026 by the authors.

Licensee MDPI, Basel, Switzerland.

This article is an open access article distributed under the terms and

conditions of the [Creative Commons](https://creativecommons.org/licenses/by/4.0/)

[Attribution \(CC BY\)](https://creativecommons.org/licenses/by/4.0/) license.

Keywords: hydrogen energy infrastructure; rolling contact fatigue; hydrogen-enhanced localized plasticity (HELP); grain refinement; subsurface cracking; spalling; delamination wear; comparative durability assessment; hydrogen embrittlement; wear indicators

1. Introduction

The development of infrastructure associated with hydrogen and hydrogen-containing fuels for internal combustion engines, including hydrogen compressors, pipeline fittings, and highly loaded tribological components, requires materials capable of withstanding the combined effects of hydrogen exposure, mechanical stresses, and rolling or rolling-sliding contact. In addition to hydrogen embrittlement [1,2], increasing attention has recently been focused on hydrogen-assisted wear and rolling contact fatigue (RCF). Under rolling contact conditions, hydrogen accelerates subsurface crack initiation and propagation, promotes premature spalling, and reduces the rolling contact fatigue life of contacting components [3–6].

Among the available corrosion-resistant structural materials, high-nitrogen Cr–Mn–N austenitic steels are of particular interest because they combine high yield strength, excellent corrosion resistance, stable non-magnetic properties, and reduced nickel content compared with conventional nickel-alloyed austenitic stainless steels. The partial replacement of nickel by manganese and nitrogen not only reduces material cost but also provides enhanced mechanical performance while maintaining a stable austenitic microstructure [7]. These characteristics make high-nitrogen Cr–Mn–N steels promising candidates for highly loaded tribological components potentially operating in hydrogen-containing environments.

The interaction of hydrogen with steels involves its diffusion through the lattice and trapping at microstructural defects, including dislocations, grain boundaries, deformation twins, carbides, nitrides, and non-metallic inclusions [8]. Hydrogen trapping sites differ significantly in their binding strength. Reversible (weak) traps, including dislocations and coherent twin boundaries, can exchange hydrogen with the surrounding lattice and therefore strongly influence the kinetics of diffusible hydrogen. In contrast, irreversible (strong) traps, such as nitrides, oxides, and other non-metallic inclusions, retain hydrogen more strongly and reduce its mobility, thereby affecting hydrogen redistribution within the microstructure [8]. Localized hydrogen accumulation at microstructural trapping sites may contribute to hydrogen-enhanced localized plasticity (HELP) and hydrogen-enhanced decohesion (HEDE), thereby facilitating crack initiation and propagation in highly stressed regions [9–12]. Hydrogen-assisted degradation is commonly interpreted through several interacting mechanisms, among which HELP and HEDE are particularly important. According to the HELP mechanism, hydrogen enhances dislocation mobility and promotes localized plastic deformation and slip localization [9,10]. In contrast, the HEDE mechanism is associated with a hydrogen-induced reduction in cohesive strength, thereby facilitating crack initiation and propagation along susceptible interfaces [10,11].

The hydrogen resistance and tribological performance of these steels strongly depend on their microstructural state, particularly grain size, deformation-twin density, dislocation substructure, and the distribution of hydrogen trapping sites [7,12,13]. Hydrogen charging can modify the wear and damage behavior of high-nitrogen steels under dry sliding and rolling contact conditions, promoting localized plastic deformation, microcrack initiation, and progressive damage within the near-surface layers [14–16].

A distinct research direction is associated with degradation diagnostics based on wear debris parameters, crack morphology, and the evolution of subsurface structures. Wear-debris characteristics can provide valuable information on the evolution of damage under rolling-sliding contact conditions [14]. In hydrogen-containing environments, such tribological observations can be considered together with hydrogen-assisted fracture and wear characteristics to provide insight into the localization and progression of material degradation [11,15,16].

Despite extensive research on hydrogen embrittlement and the tribology of steels, studies integrating hydrogen charging, microstructural state, and rolling-sliding wear

behavior in high-nitrogen austenitic steels remain limited. Consequently, an integrated methodology combining hydrogen uptake, microstructural characteristics, rolling–sliding wear behavior, and damage evolution for the comparative durability assessment of high-nitrogen austenitic steels under hydrogen exposure remains insufficiently developed.

The objective of this study is to investigate the resistance of high-nitrogen austenitic steels to hydrogen-assisted degradation and the influence of microstructural state on their rolling–sliding wear behavior, and to propose an indicator-based comparative framework for assessing the relative durability of hydrogen-charged steels based on experimentally determined wear characteristics.

The scientific hypothesis of this study is that grain refinement in cold-worked high-nitrogen Cr–Mn–N steels increases the density of grain boundaries and associated hydrogen trapping sites, while the deformation-induced defect substructure provides additional sites for hydrogen trapping, thereby promoting greater hydrogen retention and HELP-assisted localization of plastic deformation. Under rolling–sliding contact conditions, the resulting trap-assisted localization of deformation is hypothesized to promote subsurface microcrack initiation and subsequent delamination and spalling, thereby producing a more pronounced hydrogen-induced increase in wear intensity in the finer-grained steel.

The present study extends our previous investigations of the dry-friction wear behavior, hydrogen-assisted fracture, and rolling-friction behavior of high-nitrogen steels [15–17]. Some experimental wear data corresponding to the intermediate loads of 300, 400, and 500 N were reported in our previous publications, whereas the experimental results obtained at the boundary loads of 200 and 600 N are presented here for the first time. In the present work, the previously reported intermediate-load data and the new experimental results are integrated into a new combined comparative representation covering the complete investigated load range of 200–600 N; this combined representation has not been published previously. The present study further integrates wear behavior with hydrogen uptake, microstructural characteristics, grain-size statistics, Hertzian contact conditions, subsurface damage observations, statistical evaluation, and an indicator-based comparative durability assessment. Thus, the novelty of the present work lies both in the previously unpublished experimental data at 200 and 600 N and in the integrated quantitative and mechanistic comparison of the hydrogen-assisted degradation behavior of DDT68 and P900 steels under dry rolling–sliding contact conditions.

2. Literature Survey: State-of-the-Art

2.1. Hydrogen Interaction with Austenitic and High-Nitrogen Steels

Hydrogen penetration into austenitic and high-nitrogen steels occurs through interstitial diffusion, followed by trapping at microstructural defects with different binding energies, including dislocations, grain boundaries, deformation twins, carbides, nitrides, non-metallic inclusions, and regions of high stress concentration [8]. Although hydrogen diffusivity in austenitic steels is significantly lower than in ferritic steels, hydrogen trapping at microstructural defects can produce locally elevated hydrogen concentrations and modify the deformation behavior of the material [7,8]. Under appropriate combinations of hydrogen concentration and applied stress, these interactions can promote hydrogen-enhanced localized plasticity (HELP) and may also contribute to hydrogen-enhanced decohesion (HEDE), thereby facilitating crack initiation and propagation under mechanical and contact loading [9,10].

In high-nitrogen austenitic steels of the Cr–Mn–N and Fe–Mn–N systems, hydrogen can develop concentration gradients within near-surface regions and intensify planar slip and deformation localization through the hydrogen-enhanced localized plasticity (HELP) mechanism [8,12,13]. This process promotes dense slip-band formation, interactions

between slip bands and deformation twins, and increased local stress concentrations at grain boundaries and other microstructural barriers, thereby facilitating microcrack nucleation [9,12]. Under rolling contact conditions, hydrogen-assisted localized plasticity, acting in conjunction with high local contact and shear stresses, can further promote crack initiation and subsequent damage development within the near-surface region of high-nitrogen steels [15].

Hydrogen segregation at grain boundaries can reduce cohesive strength and facilitate intergranular crack initiation [10,18], whereas hydrogen accumulation near deformation twins can enhance localized dislocation activity and promote deformation localization through the hydrogen-enhanced localized plasticity (HELP) mechanism [9,12,19].

In austenitic and high-nitrogen steels, deformation twins play an important role in hydrogen-induced plasticity localization and damage initiation [12]. Within the HELP framework, hydrogen–dislocation interactions can enhance dislocation mobility and promote localized plastic deformation [9,19]. In twinning-prone austenitic steels, this enhanced dislocation activity can intensify planar slip and dislocation–twin boundary interactions, thereby promoting localized deformation and facilitating damage initiation [12].

In high-nitrogen Cr–Mn–N austenitic steels, deformation twins and other microstructural interfaces can contribute to hydrogen localization and deformation heterogeneity. Their interactions with grain boundaries and slip bands can create regions of localized stress and plastic strain, which can facilitate hydrogen-assisted microcrack initiation [12,20].

Although nitrogen decreases the hydrogen diffusion coefficient and contributes to austenite stabilization, thereby limiting hydrogen transport, it cannot completely prevent hydrogen accumulation at deformation twins and other microstructural defects [7,21]. Hydrogen accumulation at such sites can enhance localized plastic deformation and facilitate hydrogen-assisted crack initiation, even in nitrogen-alloyed steels [12,18,22].

Thus, within the near-surface layers of austenitic and high-nitrogen steels, the combined effects of hydrogen concentration gradients, elevated contact and shear stresses, and hydrogen-induced plasticity localization associated with the HELP mechanism can create favorable conditions for damage accumulation within slip bands and at twin and grain boundaries [9,12,15]. These microstructurally localized regions can become preferential sites for subsurface microcrack initiation and subsequent delamination and rolling contact fatigue damage, thereby contributing to accelerated wear and the early stages of hydrogen-assisted degradation under rolling–sliding contact conditions [3,6,15].

It should be emphasized that hydrogen-assisted degradation in high-nitrogen austenitic steels cannot be attributed exclusively to the HELP mechanism [9,10]. Owing to the high density of deformation twins, grain boundaries, non-metallic inclusions, and other microstructural heterogeneities generated during cold working, hydrogen-enhanced decohesion (HEDE) may additionally contribute to crack initiation by reducing the cohesive strength of interfaces subjected to local tensile stresses [10,18]. Furthermore, hydrogen interactions with vacancy-type defects and other microstructural heterogeneities may contribute to nanoscale defect accumulation and the formation of nanovoids, thereby providing additional sites for subsequent microcrack nucleation [11,23]. Therefore, hydrogen-assisted degradation can be interpreted as the combined action of HELP, HEDE, and hydrogen–defect interaction mechanisms, whose relative contributions depend on the local microstructure, hydrogen concentration, and stress state [10,11,23].

2.2. Microstructural Effects on Hydrogen Resistance

Grain size is one of the principal microstructural parameters governing hydrogen transport, hydrogen trapping, and the localization of plastic deformation in austenitic steels. Grain refinement increases the density of grain boundaries and consequently modifies both

hydrogen trapping and the spatial distribution of plastic deformation within hydrogen-affected regions. Depending on the alloy composition, deformation mechanism, and hydrogen-charging conditions, these changes may either promote a more homogeneous distribution of plastic strain or increase local hydrogen accumulation at microstructural trapping sites. Consequently, the balance between hydrogen trapping and plastic-strain accommodation is altered, thereby affecting susceptibility to hydrogen-assisted damage. Experimental investigations on AISI 304 austenitic stainless steel have demonstrated that grain refinement can improve resistance to hydrogen embrittlement by reducing strain localization and delaying crack initiation [24,25]. However, recent studies have shown that the effect of grain refinement is not universal and strongly depends on hydrogen-charging conditions, alloy composition, and microstructural state. In particular, an adverse effect of grain refinement on hydrogen embrittlement has been reported for high-Mn austenitic steel [26], whereas nitrogen addition combined with microstructural refinement can improve hydrogen compatibility by modifying hydrogen transport and deformation behavior in nitrogen-alloyed austenitic steels [27,28]. Grain refinement has also been shown to influence hydrogen-induced delayed fracture behavior in 316L austenitic stainless steel [29].

In high-nitrogen austenitic steels, interstitial nitrogen modifies the dislocation response through its interaction with the austenitic lattice and dislocation substructure. These interactions affect dislocation mobility and the tendency toward localized planar deformation, thereby modifying the conditions under which hydrogen-assisted localized plasticity may develop in nitrogen-alloyed austenitic steels [7,9,28].

The stability of the γ -austenite phase and the characteristics of the defect substructure, particularly planar slip, deformation twinning, and dislocation arrangements, play an important role in hydrogen transport and hydrogen-assisted degradation. Nitrogen alloying contributes to austenite stabilization and can decrease hydrogen diffusivity, thereby limiting hydrogen transport in austenitic steels [7,21,27]. Consequently, nitrogen-containing austenitic steels can exhibit improved resistance to hydrogen-induced degradation, although the magnitude of this effect depends on alloy composition and microstructural state [21,27,28]. However, recent fatigue studies have demonstrated that hydrogen can still significantly affect fatigue life, crack initiation, and fracture morphology in nitrogen-containing steels subjected to cyclic loading [22]. In Cr-Mn-N and other nitrogen-alloyed austenitic steels, the interplay between nitrogen addition, austenite stability, hydrogen trapping, and localized plastic deformation therefore plays an important role in determining susceptibility to hydrogen-assisted damage under mechanical loading [20,28].

Non-metallic inclusions, together with nitride and carbonitride precipitates, can act as preferential hydrogen trapping sites and local stress concentrators [30,31]. Under cyclic and contact loading, hydrogen localization and stress concentration associated with these microstructural heterogeneities can contribute to localized deformation and facilitate micro-crack initiation. A comprehensive analysis of hydrogen trapping mechanisms in austenitic, duplex, and supermartensitic stainless steels, including the role of inclusions, precipitates, and interfaces with different hydrogen binding energies, was presented by Silverstein and Eliezer based on thermal desorption spectroscopy investigations [30]. More recently, Qin et al. demonstrated that hydrogen trapping at non-metallic inclusions is strongly influenced by their chemical composition, size, morphology, and elastic interaction with the surrounding matrix, highlighting the importance of inclusion characteristics in controlling hydrogen localization and hydrogen-assisted degradation in steels [31].

Thus, the microstructural parameters of high-nitrogen austenitic steels—including grain size, austenite stability, nitrogen content, and the presence of inclusions and nitrides that can act as hydrogen trapping sites—govern hydrogen distribution and the localization

of deformation within near-surface regions [20,21,30,31]. Collectively, these factors establish microstructural conditions that can facilitate hydrogen-assisted microcrack initiation and subsequent delamination and rolling contact fatigue damage, thereby influencing wear resistance and relative durability under rolling–sliding contact conditions [3,6,15]. These relationships provide the basis for considering hydrogen-assisted wear and rolling contact fatigue mechanisms in the following subsection.

2.3. Hydrogen-Assisted Wear and Contact Fatigue

Under dry rolling–sliding conditions in a roller–on–roller configuration, wear evolution and subsurface damage are governed by the combined effects of contact stresses, relative sliding (slip), and cyclic plastic deformation [3,6,14,15]. Cathodic hydrogen pre-charging introduces diffusible hydrogen into the near-surface region, where it can promote localized plastic deformation and modify the mechanisms of damage initiation and propagation within the contact zone [8,9,15]. Since electrochemical hydrogen charging may itself alter the initial condition of the surface and near-surface layers, the charging procedure should be carefully controlled when interpreting rolling contact fatigue and wear results. Recent studies have demonstrated that certain electrochemical hydrogen-charging conditions may induce subsurface cracking even before subsequent mechanical testing, emphasizing the importance of distinguishing charging-induced damage from hydrogen-assisted damage developed during tribological loading [32].

Under rolling–sliding contact conditions, damage initiation frequently occurs within subsurface regions subjected to high cyclic shear stresses. In the presence of hydrogen, enhanced dislocation mobility and localized plastic deformation associated with the hydrogen-enhanced localized plasticity (HELP) mechanism can facilitate subsurface microcrack initiation within highly deformed regions and at microstructural barriers such as grain and twin boundaries and non-metallic inclusions [3,9,12]. The subsequent propagation and coalescence of such microcracks can promote delamination, material spalling, and progressive rolling contact fatigue damage [3,6,33]. Experimental investigations on bearing steels have demonstrated that hydrogen can accelerate microstructural degradation, subsurface crack evolution, and surface-originated flaking during rolling contact fatigue, resulting in a substantial reduction in fatigue life [3–6,33]. Recent investigations of bearing steels have further associated rolling contact fatigue damage with white etching crack formation and intense shear localization, providing additional insight into the microstructural evolution accompanying severe rolling contact damage [34,35]. Although comparable rolling-contact studies on high-nitrogen austenitic steels remain limited, the susceptibility of these materials to hydrogen-assisted damage is expected to be strongly influenced by austenite stability, nitrogen-dependent hydrogen transport, and the distribution and strength of hydrogen trapping sites associated with grain boundaries, dislocations, inclusions, and other microstructural defects [8,20,28,30,31].

For the assessment of relative durability under dry rolling–sliding contact conditions, it is essential to consider the pronounced influence of hydrogen on rolling contact fatigue (RCF). Classical experimental studies demonstrated that the RCF life of bearing steels decreases with increasing hydrogen content, as reported for AISI 52100 and 440C bearing steels [4]. Subsequent investigations further confirmed that hydrogen-containing environments accelerate rolling-contact damage and reduce fatigue life compared with testing in air, particularly under the high contact pressures characteristic of rolling contacts [5,6]. More recent studies have shown that hydrogen promotes microstructural degradation, subsurface crack initiation and propagation, and surface-originated flaking, ultimately contributing to premature rolling contact fatigue failure [3,33]. White etching cracks and white etching areas have also been observed during rolling contact fatigue in carburized

bearing steels and, more recently, in high-nitrogen bearing steel, demonstrating the close relationship between localized deformation, microstructural transformation, and crack evolution under severe cyclic contact loading [34,36]. Recent investigations have further associated the formation of white etching regions with severe cyclic plastic deformation and localized microstructural refinement [34–36]. In hydrogen-precharged high-nitrogen bearing steel, white etching area formation has been linked to a hydrogen-activated severe plastic deformation process accompanied by continuous dynamic recrystallization and the formation of characteristic nanoferritic grains [36]. These observations indicate that hydrogen can promote localized microstructural transformation and crack evolution under rolling contact fatigue conditions, thereby contributing to premature damage and reduced durability of rolling-contact components [33,36].

In this context, dry rolling–sliding experiments performed in a roller-on–roller configuration with prior electrolytic hydrogen charging of one contacting body provide a suitable approach for evaluating the influence of hydrogen on subsurface microcrack initiation, wear debris formation, and progressive rolling contact damage [3,14,15,32,33]. Such experiments enable hydrogen-assisted changes in wear and damage evolution to be related to the microstructural state of the contacting material, thereby establishing an experimental basis for comparative durability assessment under controlled rolling–sliding conditions.

Analysis of damage morphology and wear debris under such conditions provides a basis for identifying characteristic features of hydrogen-assisted degradation and for relating the evolution of tribological damage to comparative durability indicators under controlled rolling–sliding conditions [14–16]. This combined analysis therefore represents a useful approach for the diagnostic assessment of hydrogen-assisted wear and damage evolution, as discussed in the following subsection.

2.4. Wear-Based Diagnostics and Lifetime Assessment

For dry rolling–sliding contacts, wear intensity, expressed through mass loss, linear wear, or wear rate, provides an integral measure of material degradation during tribological loading [37,38]. Classical wear models, particularly the Archard model, relate wear volume to the applied normal load, sliding distance, and material hardness and therefore provide a fundamental framework for comparing wear behavior under conventional sliding conditions [37]. However, direct application of such relationships to rolling contact fatigue (RCF) is limited because damage evolution under cyclic rolling–sliding contact involves not only progressive material removal but also the initiation, propagation, and coalescence of surface and subsurface fatigue cracks [39–41]. Therefore, rolling-contact diagnostics should consider not only accumulated wear but also its evolution with loading cycles and the accompanying changes in damage morphology and wear debris [39,42]. An increase in wear rate associated with subsurface crack development, delamination, and spalling can indicate a transition from relatively stable material removal toward fatigue-dominated damage accumulation [39,41–43].

It should be noted that the classical Archard wear law was developed primarily for sliding wear, relating material loss to normal load, sliding distance, and material hardness [37,44]. Under rolling contact fatigue conditions, however, material loss may additionally result from cyclic surface and subsurface damage involving crack initiation and propagation, crack coalescence, delamination, and spalling [39–41]. Therefore, in the present study, the Archard framework is used only as a comparative engineering reference for interpreting wear intensity rather than as a complete physical description of hydrogen-assisted rolling contact fatigue. Accordingly, the measured wear characteristics should be interpreted together with the observed subsurface damage evolution and crack-development mechanisms.

The morphology of subsurface cracks, including their orientation, initiation depth, and relationship with regions of high cyclic shear stress, together with wear-debris characteristics such as particle size and morphology, provides important information on the dominant damage mechanisms operating during rolling–sliding contact [39–42]. This concept is consistent with the classical delamination theory of wear, according to which subsurface crack initiation, propagation, and coalescence can precede the detachment of surface material and the generation of wear debris [41]. **For rolling contact fatigue (RCF), the combined analysis of wear evolution, subsurface damage, crack morphology, and wear-debris characteristics can therefore provide complementary diagnostic information for assessing damage progression and durability in tribological components [42,45].** Hydrogen uptake in bearing steels under tribo-mechanical loading represents an additional factor that should be considered when evaluating rolling-contact degradation [46]. Microstructural investigations of bearing steels have further demonstrated that rolling contact fatigue is accompanied by the evolution of dark etching regions, white etching bands, white etching areas and white etching cracks, and other localized microstructural alterations associated with severe cyclic deformation and crack development [34,35,47,48]. Hydrogen can further accelerate the formation and evolution of such microstructural alterations under rolling contact fatigue conditions [43,49]. Accordingly, wear-debris analysis, when interpreted together with subsurface microstructural observations and wear evolution, can contribute to identifying transitions from relatively stable wear toward fatigue-dominated damage and premature material removal [14,42,48].

At the methodological level, wear-debris-based diagnostics provides a useful approach for assessing the condition of tribological systems through characterization of particle features, including their number, size distribution, morphology, composition, and evolution during operation [14,42]. When combined with information on wear evolution and subsurface damage, wear-debris characteristics can provide complementary information on the progression of rolling-contact degradation [39,42,48]. In rolling contact fatigue, subsurface crack development and microstructural alterations, including white etching areas and white etching cracks, are associated with severe cyclic deformation, while hydrogen can further promote their formation and evolution under hydrogen-assisted conditions [34–36,43,49]. Therefore, correlating wear-debris characteristics with subsurface microstructural observations and damage evolution represents a promising approach for identifying hydrogen-assisted transitions from relatively stable wear toward fatigue-dominated material removal.

3. Materials and Methods

Products of wear generated from cold-deformed high-nitrogen manganese steels were investigated. The rolling–sliding wear tests were performed using a modified 2070 SMT-1 friction testing machine, which was modernized at the Karpenko Physico-Mechanical Institute of the National Academy of Sciences of Ukraine (Lviv, Ukraine). The rolling–sliding mode was established by repositioning the drive belts between the standard fixed pulleys of the machine, thereby changing the transmission ratio between the roller shafts. The nominal rotational speeds of the lower and upper roller drive systems were 1480 rpm and 1240 rpm, respectively. The lower roller was manufactured from quenched Steel 45 (diameter 42 mm), whereas the upper rollers were manufactured from DDT68 (diameter 35 mm) and P900 (diameter 31 mm). Their chemical compositions and properties are presented in Tables 1 and 2. Because different upper-roller diameters were employed for DDT68 and P900 steels, the corresponding Hertzian contact parameters were calculated separately for each roller configuration, as presented in Table 3.

Table 1. Chemical composition of the studied steels (wt.%).

Index	Brand	C	N	V	Mo	Ni	Cr	Mn	Si
1	Steel DDT 68	0.06	0.97	0.14	2.08	0.13	17.5	19.4	0.52
2	Steel P900	0.08	0.58	0.12	0.13	1.18	17.8	19.0	0.38
3	Steel 45 *	0.47	–	–	–	0.23	0.24	0.78	0.27

* up to 0.04 wt.% S, up to 0.035 wt.% P, up to 0.25 wt.% C, up to 0.25 wt.% Cu, up to 0.08 wt.% As.

Table 2. Properties of high nitrogen steels.

№	Steel	σ_B , MPa	σ_2 , MPa	δ , %	ψ , %	K_{1c} , MPa $\sqrt{m}$	HRc
1	Steel DDT 68	658	1046	62	66	397	52
2	Steel P900	560	921	67	66	535	48

Table 3. Calculated Hertzian contact parameters for the investigated roller pairs.

Applied Load, N	pmax, GPa DDT68–Steel 45	Depth of τ_{max} , mm DDT68–Steel 45	pmax, GPa P900–Steel 45	Depth of τ_{max} , mm P900–Steel 45	τ_{max} , GPa DDT68–Steel 45	τ_{max} , GPa P900–Steel 45
200	0.277	0.036	0.287	0.035	0.083	0.086
300	0.340	0.044	0.352	0.043	0.102	0.105
400	0.392	0.051	0.406	0.049	0.118	0.122
500	0.439	0.057	0.454	0.055	0.132	0.136
600	0.480	0.062	0.497	0.060	0.144	0.149

The microstructures of the alloys consisted of an austenitic metallic matrix with a microhardness of 4.2–5.0 GPa. The experiments were performed under dry rolling–sliding contact at normal loads ranging from 200 to 600 N.

Fractographic investigations were performed using an EVO-40XVP electron microscope (Carl Zeiss SMT AG, Oberkochen, Germany) equipped with an INCA Energy 350 microanalysis system (Oxford Instruments Plc, Abingdon, UK). The microstructure was investigated by Zeiss Stemi 2000-C Stereo Microscopes (Carl Zeiss AG, Oberkochen, Germany) and SIGETA industrial color digital camera (UCMOS 1300, 1.3 MP and SIGETA International Color Digital Camera MCMOS 5100 5.1 MP) (SIGETA, Hong Kong, China).

Each tribological experiment was performed independently three times under identical testing conditions. The reported wear intensity values represent the arithmetic mean of three independent measurements and are presented as mean $\pm$ standard deviation (SD). Error bars in all wear-intensity plots represent one standard deviation. Statistical significance of the differences between DDT68 and P900 steels was evaluated using the two-tailed Student's *t*-test. Differences were considered statistically significant at $p < 0.05$. Polynomial regression was employed exclusively to visualize the experimental trends and facilitate comparison between the investigated materials. The statistical significance of the observed differences was evaluated independently using Student's *t*-test and was not inferred from the regression coefficients (R^2).

To facilitate comparison with conventional rolling contact fatigue studies, the elastic contact parameters were calculated using the classical Hertzian solution for line contact between two parallel cylindrical rollers. The calculations were performed separately for each roller pair using a contact width of 10 mm, Young's modulus $E = 210$ GPa, and Poisson's ratio $\nu = 0.30$ for both contacting steels. The maximum Hertzian contact pressure p_{max} , maximum subsurface shear stress τ_{max} , and the corresponding depth below the contact surface were determined for each applied load and roller pair. These calculations account for the slight differences in roller geometry between the DDT68 and P900 configurations and provide the basis for the comparative interpretation of the tribological results presented below. The calculated values are summarized in Table 3.

The calculated maximum Hertzian contact pressures ranged from 0.277 to 0.480 GPa for the DDT68–Steel 45 pair and from 0.287 to 0.497 GPa for the P900–Steel 45 pair. The corresponding maximum subsurface shear stresses were located approximately 0.035–0.062 mm below the contact surface. These depths define the near-surface region in which cyclic plastic deformation and hydrogen-assisted subsurface crack initiation are expected to develop.

High-nitrogen austenitic Cr–Mn stainless steels exhibit a unique combination of mechanical properties, including strength, toughness, corrosion resistance, and non-magnetic behavior. Nitrogen is highly effective in increasing the strength of austenitic stainless steels without reducing their ductility. This behavior is observed at nitrogen levels of approximately 1%, which are achieved through the electroslag remelting process.

The steels were produced from electrodes melted in an arc furnace, which were subsequently remelted in a pressurized electroslag remelting unit to improve purity. After machining the billets, the samples underwent diffusion annealing followed by cold working. They were then annealed to relieve stress and subsequently subjected to final finishing.

Grain-size statistics were evaluated using the arithmetic mean, standard deviation, variance, and coefficient of variation (CV), calculated according to standard descriptive statistical procedures. These statistical parameters were used to characterize the dispersion and heterogeneity of the grain-size distributions in the investigated steels. Grain-size statistics were based on measurements of approximately 350 individual grains for each material condition, providing a statistically representative characterization of the microstructure. The calculated statistical parameters are summarized in Table 4.

Table 4. Statistical processing of microstructure parameters.

Index	Indicators	Brand	
		DDT68	P900
1.	Maximum	87	132
2.	Minimum	13.01	23.43
3.	Range of variation	73.98	106.57
4.	Average linear deviation	12.62	22.86
5.	Population variance	278.34	823.36
6.	Sample variance	278.93	826.05
7.	Standard deviation	16.68	28.69
8.	Sample standard deviation	16.70	28.74
9.	Coefficient of variation	67%	64%
10.	Oscillation coefficient	2.14	2.49

Mathematical processing of the results and the construction of histograms were performed using Microsoft Excel and OriginLab software 2021b.

The specimens were cathodically charged in a 26 wt.% H₂SO₄ solution containing 5 mg/L As₂O₃ for 2 h at 25 °C. For the comparative tribological experiments, a current density of 0.5 A/cm² was selected, and all specimens were charged under identical conditions to ensure a consistent and reproducible hydrogen charging procedure. The electrolyte composition was maintained throughout the charging process without replacement. To minimize the loss of diffusible hydrogen, all rolling–sliding wear tests were initiated immediately after completion of the hydrogen charging procedure. The time interval between hydrogen charging and tribological testing was kept constant for all specimens to ensure the reproducibility and comparability of the experimental results.

Hydrogen concentration was determined using rectangular specimens (5 × 10 × 50 mm) manufactured from P900 and DDT68 steels. These specimens were hydrogen charged under the same electrochemical conditions as the tribological rollers

(electrolyte composition, current density, charging duration, and temperature). Consequently, the measured hydrogen concentrations represent the characteristic hydrogen uptake achieved under the selected charging conditions.

Following electrochemical hydrogen charging, the witness specimens were transferred immediately to the LECO hydrogen analyzer to minimize the loss of diffusible hydrogen prior to measurement. No separate degassing procedure was applied before hydrogen extraction. The hydrogen content was determined by the vacuum-fusion method using a LECO TCH 600 analyzer [50].

4. Results and Discussions

4.1. Microstructural Parameters of the Investigated Steels

The microstructure of the investigated high-nitrogen steels consisted predominantly of an austenitic matrix containing non-metallic inclusions, which appeared as dark features in the metallographic microsections (Figure 1).

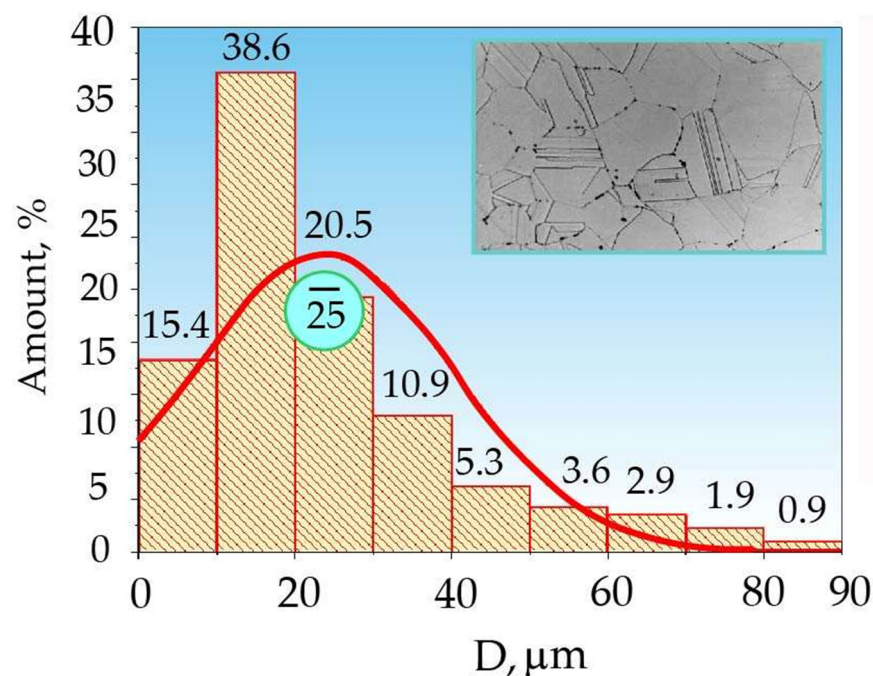


Figure 1. Microstructure of the high-nitrogen steel DDT68 and histogram of grain size distribution (average grain diameter, D).

DDT68 steel (Figure 1) exhibited a relatively fine-grained austenitic microstructure with clearly defined grain boundaries and deformation twins visible within several grains. Dark features, attributed to non-metallic inclusions, were also observed within the microstructure, including regions adjacent to grain boundaries. The presence of deformation twins is consistent with the cold-worked state of the investigated steel and indicates substantial deformation-induced modification of the austenitic microstructure.

In Figure 2, the microstructure of the high-nitrogen austenitic steel P900 is shown, which differs substantially from that of DDT68 presented in Figure 1. The austenite grains exhibit a noticeably larger size with well-defined and smooth grain boundaries. The amount of defects—specifically inclusions, pores, and deformation bands—is significantly reduced. Residual twins are observed in some grains; however, their intensity is markedly lower compared with the microstructure of DDT68 shown in Figure 1.

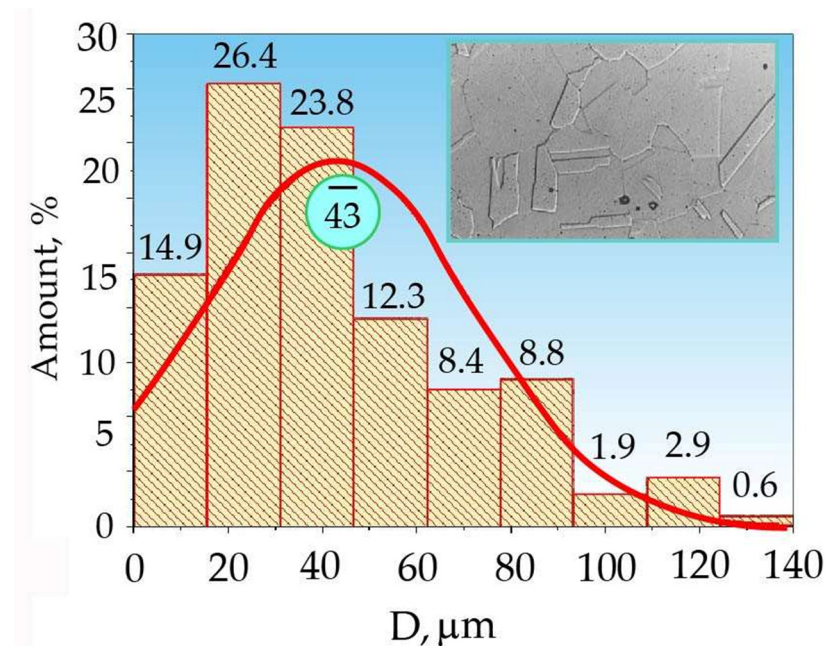


Figure 2. Microstructure of the high-nitrogen steel P900 and histogram of grain size distribution (average grain diameter, D).

Overall, the microstructures shown in Figures 1 and 2 exhibit predominantly austenitic matrices with differences in grain size, grain morphology, and the distribution of deformation-induced features. DDT68 is characterized by a finer-grained structure, whereas P900 exhibits a comparatively coarser-grained morphology. Deformation twins and non-metallic inclusions are observed within the investigated microstructures, although their distribution differs between the two steels. These microstructural differences reflect the distinct structural states of DDT68 and P900 and may influence their subsequent hydrogen uptake and deformation behavior under rolling–sliding contact conditions.

Analysis of the grain-size distributions shown in Figures 1 and 2 demonstrates a pronounced difference between the investigated steels. Based on the quantitative measurements of 350 grains for each steel, the mean grain size was approximately 25 μm for DDT68 and 43 μm for P900. Thus, DDT68 is characterized by a substantially finer-grained microstructure than P900. The fitted distribution curves shown in Figures 1 and 2 are provided to illustrate the corresponding grain-size distributions. Statistical analysis of the measured grain-size distributions was performed for both steels, and the resulting descriptive statistical parameters are presented in Table 4.

The calculated coefficients of variation for both steels exceed 33%, indicating broad grain-size distributions and substantial microstructural heterogeneity. This observation is consistent with the heterogeneous microstructure characteristic of cold-worked high-nitrogen austenitic steels. The observed variability is attributed to heterogeneous grain refinement, localized deformation bands, deformation twins, and non-uniform dislocation substructures introduced during cold working, which produce spatial variations in potential hydrogen-trapping sites. Therefore, the statistical dispersion reported in Table 4 represents an inherent characteristic of the investigated microstructures rather than a limitation of the experimental methodology. Consequently, the conclusions of the present study are based on the combined interpretation of grain-size statistics, hydrogen concentration measurements, tribological performance, wear-debris morphology, and SEM observations of subsurface damage evolution, rather than on grain-size statistics alone.

Hydrogen Charging of the High-Nitrogen Steels

Following cathodic hydrogen charging, both investigated high-nitrogen austenitic steels exhibited measurable hydrogen uptake. The hydrogen concentration determined for DDT68 was 15–17 ppm, whereas that measured for P900 was 12–14 ppm. Thus, under identical hydrogen-charging conditions, the finer-grained DDT68 steel exhibited a higher hydrogen uptake than the coarser-grained P900 steel. This difference is consistent with the influence of microstructural state on hydrogen trapping and retention, since grain refinement increases the density of grain boundaries and, together with dislocations, deformation twins, precipitates, and other microstructural defects introduced or modified by cold working, changes the population and distribution of potential hydrogen trapping sites [8,24–31]. However, the measured difference in hydrogen content cannot be attributed exclusively to grain size, because the two steels also differ in chemical composition, including their nitrogen and molybdenum contents, as well as in their deformation-induced defect structures. Therefore, the higher hydrogen concentration measured in DDT68 is interpreted as the combined result of its chemical composition and microstructural state rather than as evidence of a single dominant trapping mechanism.

The hydrogen concentrations measured after charging were: DDT68, 15–17 ppm; P900, 12–14 ppm.

The reported hydrogen concentrations represent the experimentally observed minimum-to-maximum ranges obtained under identical hydrogen-charging and measurement conditions. These values do not represent mean concentrations with associated standard deviations and were used exclusively for comparative evaluation of the two steels under the same experimental conditions. Accordingly, the reported hydrogen concentration ranges should be interpreted within the experimental scope of the present study rather than as statistically characterized population values.

The depth of the hydrogen-affected region depends strongly on the material microstructure and hydrogen-charging conditions, including current density, charging time, and electrolyte composition. For example, Helen P. Kyriakopoulou, P. Karmiris-Obratański, et al. reported hydrogen penetration depths of approximately 50 and 180 μm in X65 pipeline steel after 26 h of in situ cathodic hydrogen charging at current densities of 10 and 20 mA/cm^2 , respectively [51]. These results demonstrate that electrochemical hydrogen charging can affect a substantial near-surface region; however, the reported penetration depths should not be directly extrapolated to the DDT68 and P900 steels investigated in the present study because of differences in material composition, microstructure, and hydrogen-charging conditions. Accordingly, the hydrogen penetration depth in the investigated high-nitrogen austenitic steels was not inferred from literature data, and the experimentally measured hydrogen concentration ranges were used for their comparative assessment.

4.2. Wear Intensity Results for the Investigated Steels

Figure 3 illustrates the change in wear intensity, I (normalized per 1000 m of sliding distance), as a function of normal load, P , under dry rolling–sliding conditions prior to hydrogen charging. The combined dependencies shown in Figure 3 were newly constructed for the present study by integrating previously reported experimental data at the intermediate loads of 300, 400, and 500 N with previously unpublished results obtained at 200 and 600 N; the combined graphical representation itself has not been published previously. The plot presents four curves corresponding to the investigated friction pairs, together with second-order polynomial trend lines and their coefficients of determination (R^2). The polynomial fitting was employed exclusively as an empirical means of visualizing the nonlinear experimental trends and facilitating comparison between the investigated

material conditions; it was not intended to represent a mechanistic wear law, provide a generalized predictive wear model, or be used for extrapolation beyond the investigated load range. A power-law formulation was not adopted because the limited number of experimental load levels did not permit a physically validated comparison between alternative wear-law formulations. Accordingly, the fitted curves and R^2 values should be interpreted only as descriptive characteristics of the experimental trends within the investigated loading range.

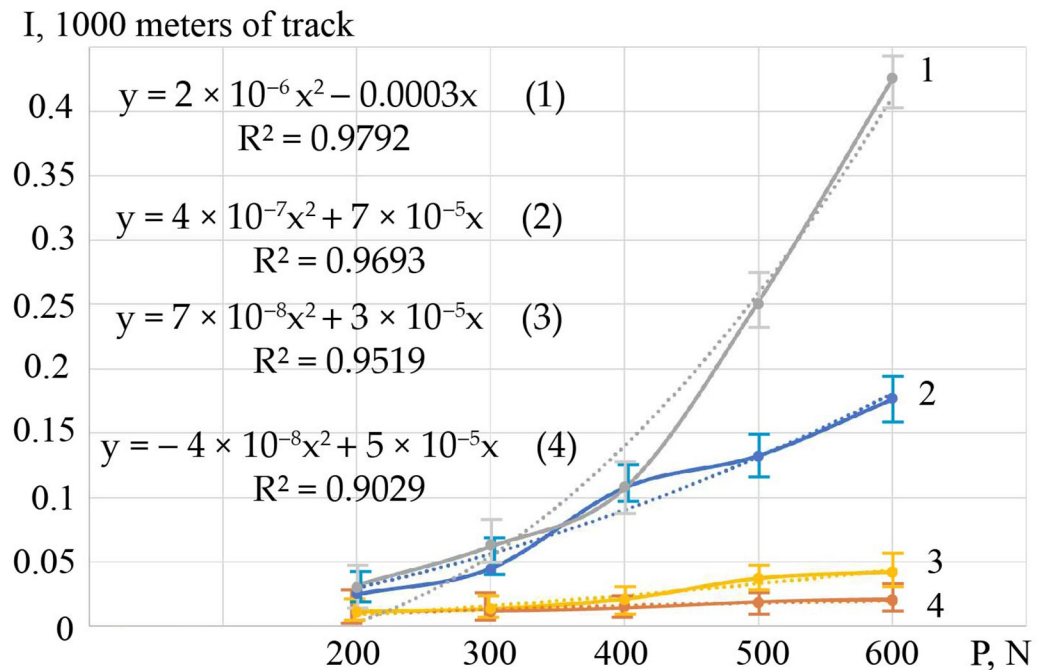


Figure 3. Wear intensity as a function of applied load under dry rolling–sliding contact conditions before hydrogen charging. Values are presented as mean $\pm$ standard deviation (SD) from three independent experiments. Error bars represent one standard deviation. 1—P900 steel; 2—DDT68 steel; 3,4—Steel 45. Friction pairs: (1,3) and (2,4). The numbering of the polynomial equations and the corresponding coefficients of determination (R^2) corresponds to the numbering of the plotted curves.

Curve 1 (P900 steel) exhibits the most pronounced increase in wear intensity with increasing normal load. From approximately 300–350 N, the wear intensity, I , increases markedly, reaching the highest value among the investigated material conditions at 600 N. The coefficient of determination ($R^2 = 0.9792$) indicates a close descriptive fit of the second-order polynomial to the experimental data within the investigated load range. As noted above, this polynomial relationship is used solely to characterize the observed experimental trend and should not be interpreted as a predictive wear model.

The pronounced increase in wear intensity observed for P900, particularly above approximately 300 N, may be associated, in part, with its comparatively coarse-grained austenitic microstructure, characterized by a mean grain size of approximately 43 μm . The lower grain-boundary density associated with the larger grain size can modify the accommodation and localization of plastic deformation under contact loading. As the normal load increases, the corresponding increase in Hertzian contact stresses promotes more intensive cyclic deformation and damage accumulation within the near-surface and subsurface regions. Consequently, the marked increase in wear intensity observed for P900 at higher loads is interpreted as reflecting the combined influence of its microstructural state and the increasing severity of rolling–sliding contact loading.

Curve 2 (DDT68 steel) generally exhibits lower wear intensity than P900 over the investigated load range of 200–600 N. At 200 N, the wear intensities of DDT68 and P900 are nearly identical; however, with increasing normal load, the wear intensity of DDT68 increases more gradually than that of P900. Consequently, the difference in wear behavior between the two steels becomes more pronounced at higher loads. The coefficient of determination ($R^2 = 0.9693$) indicates a close descriptive fit of the second-order polynomial to the experimental data within the investigated load range.

DDT68 steel (**Curve 2**), in contrast, is characterized by a finer-grained austenitic microstructure, with a mean grain size of approximately 25 μm , compared with approximately 43 μm for P900. Deformation twins and other deformation-induced microstructural features are also observed in DDT68. Grain refinement increases the density of grain boundaries, which can influence the distribution and accommodation of plastic deformation under contact loading.

Curves 3 and 4, corresponding to the wear response of quenched Steel 45 in contact with DDT68 and P900, respectively, exhibit the lowest wear-intensity values over the investigated load range. The comparatively high wear resistance of Steel 45 can be associated with its high hardness (55–62 HRC) and quenched martensitic microstructure, which provide increased resistance to plastic deformation and material removal under contact loading. Consequently, Steel 45 exhibits lower wear intensity than the investigated high-nitrogen austenitic steels under the present dry rolling–sliding conditions. The relatively small difference between **Curves 3 and 4** further indicates that changing the opposing high-nitrogen steel from DDT68 to P900 produced only a limited difference in the measured wear response of the Steel 45 counterbody within the investigated load range.

Thus, the obtained experimental results demonstrate that, under dry rolling–sliding conditions prior to hydrogen charging, P900 exhibits the highest wear intensity, whereas DDT68 demonstrates comparatively higher wear resistance over the investigated load range of 200–600 N. The lowest wear-intensity values were observed for quenched Steel 45 in both investigated friction pairs. These results indicate a clear difference in the wear response of the investigated materials under increasing normal load and provide the reference condition for subsequent evaluation of the effect of hydrogen charging on their wear behavior. Figure 3 presents a unified comparative summary of the complete dry rolling–sliding wear dataset obtained in the authors' experimental program, integrating previously reported intermediate-load data with previously unpublished results at 200 and 600 N.

Figure 4 summarizes the effect of hydrogen charging on the wear behavior of the investigated high-nitrogen steels. Figure 4a compares the experimentally determined mean wear intensities before and after hydrogen charging over the entire applied load range, whereas Figure 4b presents the hydrogen-induced increase in wear intensity expressed as the ratio I_H/I_0 . This representation enables comparison of the hydrogen-associated wear response of DDT68 and P900 steels under the investigated dry rolling–sliding contact conditions with steel 45.

Hydrogen charging increased the wear intensity of both investigated steels throughout the applied load range. However, the magnitude of this increase differed markedly between the two materials. For DDT68 steel, the wear-intensity ratio I_H/I_0 increased from approximately 2.1 at 200 N to approximately 10.0 at 600 N. In contrast, for P900 the wear-intensity ratio varied non-monotonically from approximately 1.4 to 4.2 over the investigated load range. These results indicate a substantially more pronounced hydrogen-associated increase in wear intensity for DDT68 under the investigated experimental conditions.

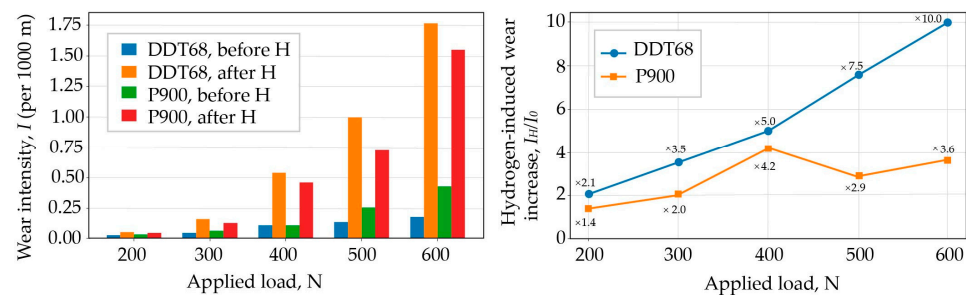


Figure 4. Comparison of the wear behavior of DDT68 and P900 steels before and after hydrogen charging under dry rolling–sliding contact with steel 45: **(left)** wear intensity as a function of the applied load; **(left)** Hydrogen-induced increase in wear intensity (I_H/I_0) as a function of applied load.

For P900 steel, the I_H/I_0 ratio exhibited a local maximum of approximately 4.2 at 400 N, followed by a decrease to approximately 2.9 at 500 N and a subsequent increase to approximately 3.6 at 600 N. This non-monotonic behavior of the relative wear ratio does not indicate a decrease in the absolute wear intensity, which continued to increase with applied load. The temporary reduction in I_H/I_0 at 500 N may reflect a greater capacity of the P900 austenitic matrix for local plastic accommodation and structural adaptation during running-in under rolling–sliding contact. In contrast, DDT68 accumulated a higher hydrogen concentration (15–17 ppm) than P900 (12–14 ppm) under identical charging conditions. The combination of higher hydrogen uptake and the finer, more strongly strengthened microstructural state of DDT68 may promote hydrogen-assisted localization of plastic deformation and limit the ability of the near-surface microstructure to accommodate increasing cyclic contact stresses through distributed plastic deformation. Thus, the different shapes of the I_H/I_0 dependencies may reflect a competition between plastic accommodation during running-in and hydrogen-assisted damage accumulation. For P900, the former mechanism may remain effective over an intermediate load range, whereas in DDT68 the higher hydrogen uptake may promote earlier dominance of localized hydrogen-assisted damage at high contact loads.

The observed differences in wear intensity between DDT68 and P900 steels after hydrogen charging exceeded the experimental scatter. Statistical analysis using the two-tailed Student's t -test confirmed statistically significant differences between the investigated steels at all considered load levels ($p < 0.05$). These results support the conclusion that the differences in hydrogen-associated wear response between DDT68 and P900 cannot be attributed solely to experimental variability.

To further support the statistical reliability of the obtained results, the mean values, standard deviations, and results of the Student t -test are summarized in Table 5.

Table 5. Statistical summary of wear intensity measurements before and after hydrogen charging under dry rolling–sliding contact conditions.

Applied Load (N)	DDT68, Before H (Mean $\pm$ SD)	DDT68, After H (Mean $\pm$ SD)	P900, Before H (Mean $\pm$ SD)	P900, After H (Mean $\pm$ SD)	Wear Increase, DDT68 (I_H/I_0)	Wear Increase, P900 (I_H/I_0)	p -Value
200	0.030 $\pm$ 0.003	0.063 $\pm$ 0.005	0.025 $\pm$ 0.003	0.035 $\pm$ 0.003	$\times 2.1$	$\times 1.4$	0.0012
300	0.050 $\pm$ 0.006	0.175 $\pm$ 0.011	0.055 $\pm$ 0.005	0.112 $\pm$ 0.012	$\times 3.5$	$\times 2.0$	0.0024
400	0.105 $\pm$ 0.004	0.527 $\pm$ 0.014	0.105 $\pm$ 0.012	0.444 $\pm$ 0.034	$\times 5.0$	$\times 4.2$	0.0167
500	0.135 $\pm$ 0.008	1.001 $\pm$ 0.026	0.250 $\pm$ 0.008	0.725 $\pm$ 0.029	$\times 7.4$	$\times 2.9$	0.00025
600	0.175 $\pm$ 0.014	1.753 $\pm$ 0.066	0.420 $\pm$ 0.018	1.523 $\pm$ 0.044	$\times 10.0$	$\times 3.6$	0.0073

Values are presented as mean $\pm$ standard deviation (SD), based on three independent tribological experiments for each testing condition ($n = 3$). Statistical significance between

DDT68 and P900 after hydrogen charging at each applied load was evaluated using a two-tailed Student's *t*-test.

The statistical analysis demonstrates that the differences between DDT68 and P900 steels remain statistically significant throughout the investigated load range. The relatively small standard deviations indicate good repeatability of the tribological measurements and support the reliability of the observed hydrogen-assisted wear behavior.

After hydrogen charging, DDT68 (blue columns) exhibits the highest wear intensity among the investigated high-nitrogen steels. With increasing normal load from 200 to 600 N, its wear intensity increases progressively, reaching the maximum value at 600 N. Compared with the behavior observed prior to hydrogen charging, this pronounced increase indicates a higher sensitivity of DDT68 to hydrogen-assisted wear degradation under the investigated rolling–sliding conditions. This behavior may be associated with its finer-grained microstructure (mean grain size $\approx 25\ \mu\text{m}$) and the correspondingly higher density of grain boundaries and other deformation-related microstructural trapping sites, which can promote hydrogen retention and localized hydrogen-assisted deformation [8,9,12,24–31]. Consistent with this interpretation, DDT68 exhibited a higher measured hydrogen concentration (15–17 ppm) than P900 (12–14 ppm) after identical hydrogen-charging conditions. Under increasing contact loading, the combined effects of hydrogen accumulation, localized plastic deformation, and cyclic contact stresses can facilitate microcrack initiation and subsequent damage accumulation, thereby contributing to the observed increase in wear intensity [3,9,12,15].

After hydrogen charging, P900 steel (grey columns) exhibits lower wear intensity than DDT68 over the investigated load range of 200–600 N. Although P900 has a comparatively coarser-grained austenitic microstructure (mean grain size $\approx 43\ \mu\text{m}$), its response to hydrogen charging is less pronounced than that of the finer-grained DDT68 steel. This difference is consistent with the lower experimentally measured hydrogen concentration in P900 (12–14 ppm) compared with DDT68 (15–17 ppm). The lower hydrogen-assisted wear response of P900 may therefore reflect the combined influence of its chemical composition, austenite stability, hydrogen transport and trapping behavior, and deformation-induced microstructural state rather than grain size alone [20,21,28,52–54].

Steel 45, in both contact pairs (orange and yellow bars), exhibits the lowest wear-intensity values among the investigated materials, with comparatively small changes over the investigated load range. Its high hardness (55–62 HRC) and quenched martensitic microstructure provide high resistance to plastic deformation and material removal under rolling–sliding contact, consistent with recent studies demonstrating the beneficial influence of martensitic microstructures and increased hardness on the wear resistance of steels under sliding and rolling–sliding conditions [55–57]. Moreover, the comparatively stable wear response of Steel 45 when paired with hydrogen-charged DDT68 and P900 indicates that hydrogen pre-charging of the opposing high-nitrogen steel had a substantially smaller effect on the wear of the Steel 45 counterbody than on the wear response of the hydrogen-charged high-nitrogen steels themselves.

Thus, hydrogen charging produces a distinctly material-dependent change in wear behavior under dry rolling–sliding conditions. DDT68 exhibits the highest susceptibility to hydrogen-assisted wear degradation, with the hydrogen-induced wear increase becoming progressively more pronounced with increasing load, whereas P900 demonstrates a substantially lower response to hydrogen charging. This difference is consistent with the higher hydrogen uptake measured for DDT68 (15–17 ppm) compared with P900 (12–14 ppm) and supports the conclusion that the tribological response to hydrogen is governed by the combined effects of hydrogen uptake, microstructural state, and applied contact loading [58]. In contrast, the comparatively low and stable wear intensity of Steel 45 indicates that

hydrogen pre-charging of the opposing high-nitrogen steel produced only a limited change in the wear response of the quenched martensitic counterbody under the investigated conditions [55–57].

The higher hydrogen-assisted wear sensitivity of DDT68 cannot be attributed to a single microstructural or compositional factor but should be interpreted as the combined consequence of grain size, deformation-induced defect structure, hydrogen trapping and uptake, alloy composition, and the applied contact stress state. DDT68 has a substantially finer austenitic microstructure than P900, with mean grain sizes of approximately 25 and 43 μm , respectively. The finer microstructure provides a larger grain-boundary area per unit volume and, together with deformation-induced defects generated by cold working, increases the population of potential hydrogen-trapping sites. Recent investigations of cold-deformed austenitic stainless steels have confirmed that deformation-induced defect structures can significantly modify hydrogen uptake and transport behavior [59]. Consistent with this interpretation, DDT68 accumulated 15–17 ppm hydrogen under identical charging conditions, compared with 12–14 ppm for P900. Thus, the experimentally observed difference in hydrogen uptake provides a direct basis for considering microstructure-dependent hydrogen trapping as one of the factors contributing to the higher hydrogen-assisted wear response of DDT68.

Under dry rolling–sliding contact, retained hydrogen interacts with the deformation substructure in regions subjected to high cyclic contact and shear stresses. Within the HELP framework, hydrogen can enhance dislocation mobility and promote localized plastic deformation rather than uniformly reducing resistance to dislocation motion [9,10]. Recent studies on austenitic stainless steels have further demonstrated that hydrogen charging can increase dislocation density, intensify deformation twinning, and modify the localization of plastic deformation, thereby facilitating crack nucleation at deformation-induced microstructural barriers [60,61]. In a cold-worked microstructure containing dislocations, deformation twins, grain boundaries, and other trapping sites, such localization can promote microcrack initiation and subsequent crack coalescence under cyclic contact loading. Previous investigations have also demonstrated a coupling between mechanically induced stresses in near-surface layers and hydrogen absorption and diffusion, although these observations were obtained under cavitation rather than rolling–sliding conditions [62]. This stress–hydrogen interaction provides additional support for considering the local contact-stress state as an important factor governing hydrogen redistribution and damage development in the near-surface region. Collectively, these mechanisms provide a plausible explanation for the pronounced increase in wear intensity observed for hydrogen-charged DDT68 with increasing normal load.

The difference between DDT68 and P900 should not be interpreted solely in terms of their nickel contents. Although P900 contains more Ni than DDT68 (approximately 1.18% and 0.13%, respectively), the steels simultaneously differ in nitrogen, molybdenum and carbon contents, grain size, and cold-worked deformation substructure. Nitrogen contributes to austenite stabilization and modifies hydrogen absorption, trapping, and transport behavior, while cold-working-induced defects provide additional hydrogen trapping sites and alter the deformation response of the austenitic matrix [27,59,63–65]. Recent investigations of nickel-economized austenitic stainless steels further demonstrate that susceptibility to hydrogen-assisted degradation depends on the combined effects of alloy composition, austenite stability, and deformation mode rather than on nickel content alone [65]. Consequently, the lower hydrogen-assisted wear response of P900 is more appropriately interpreted as the result of the combined effects of alloy composition, austenite stability, hydrogen uptake and trapping, grain size, and deformation-induced microstructure rather than as a beneficial effect of Ni alone.

The present results therefore reveal an important distinction between wear resistance in the hydrogen-free and hydrogen-charged states. Before hydrogen charging, the finer-grained DDT68 exhibited lower wear intensity than P900 over the 300–600 N load range, indicating a favorable effect of its microstructural state under conventional dry rolling–sliding conditions. After hydrogen charging, however, DDT68 exhibited a substantially greater hydrogen-induced increase in wear intensity than P900. This change in relative tribological response indicates that microstructural features contributing to favorable wear resistance in the hydrogen-free state may simultaneously increase hydrogen retention and promote localized hydrogen-assisted deformation after charging [15–17,58]. Such behavior is consistent with the reported non-universal effect of grain refinement on hydrogen resistance, which depends on the balance between grain-boundary strengthening, hydrogen trapping, deformation mechanisms, and the resulting localization of plastic strain [24–29,59]. The resulting competition between grain-refinement-related strengthening and trap-assisted hydrogen localization therefore provides a plausible explanation for the markedly greater hydrogen sensitivity of DDT68 under the present rolling–sliding conditions. Accordingly, the experimental results support the proposed hypothesis that the finer-grained, cold-worked microstructural state of DDT68, while favorable for wear resistance in the hydrogen-free condition, can promote greater hydrogen-assisted degradation after hydrogen uptake.

Thus, under the investigated experimental conditions, P900 exhibited a lower hydrogen-induced increase in wear intensity than DDT68. This difference should not be attributed to a single compositional or microstructural parameter but rather to the combined influence of hydrogen uptake, grain size, alloy composition, austenite stability, hydrogen trapping, and the cold-worked deformation substructure [27,59,63,64]. In particular, the lower hydrogen concentration measured in P900 (12–14 ppm) compared with DDT68 (15–17 ppm) is consistent with its less pronounced hydrogen-assisted wear response. These results further demonstrate that the wear resistance of high-nitrogen austenitic steels under hydrogen exposure cannot be predicted from their hydrogen-free tribological behavior alone, because hydrogen uptake and microstructure-dependent trapping can substantially modify the material response under rolling–sliding contact.

Figure 5 presents representative SEM images of one characteristic hydrogen-assisted crack observed in DDT68 steel after rolling–sliding testing following electrochemical hydrogen charging. The images correspond to different regions and magnifications of the same crack and illustrate its morphology, branching, delamination features, and crack-tip characteristics. The figure is intended to provide qualitative morphological evidence of hydrogen-assisted damage rather than to document the time-resolved evolution of crack growth or to provide a quantitative assessment of crack density.

In panel (a), a schematic representation of the roller indicates the location of the observed crack within the rolling–sliding contact zone.

The general view of the crack shown in panel (b) indicates its surface or near-surface origin within the contact zone. The surrounding morphology is consistent with localized deformation and damage developing under the combined influence of rolling–sliding contact stresses and hydrogen.

The micrographs in panels (c–f), obtained from different regions and at different magnifications of the same crack, reveal variations in crack opening, local branching, delamination, and characteristic damage features.

The central regions shown in panels (c,d) reveal a system of secondary microcracks branching from the primary crack. The observed crack paths suggest the involvement of both intergranular and transgranular fracture modes, resulting in the formation of a complex branched crack network within the near-surface region.

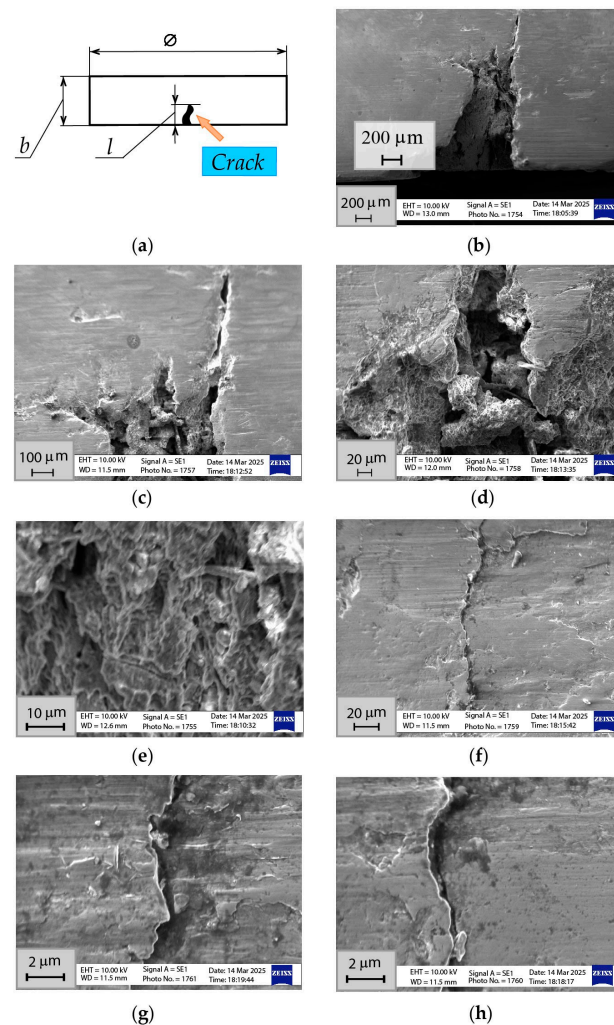


Figure 5. Representative morphology of a hydrogen-assisted crack in DDT68 steel after rolling-sliding testing following electrochemical hydrogen charging. Images (a–h) illustrate different regions and magnifications of the same characteristic crack, highlighting crack morphology, branching, delamination features, and crack-tip characteristics. Roller geometry notation: $\varnothing$ —roller diameter; b —roller width; l —crack length. (a) Overall view of the representative hydrogen-assisted crack ($\times 30$). (b) Crack-tip region ($\times 90$). (c) Central crack region ($\times 300$). (d) Internal crack region ($\times 1500$). (e) Terminal crack region ($\times 500$). (f) Terminal crack region ($\times 2000$). (g) Terminal crack region ($\times 2000$). (h) Crack attenuation zone ($\times 2000$).

In the higher-magnification images (e–h), delamination facets and relatively brittle-appearing regions are observed together with areas exhibiting more pronounced plastic deformation. The coexistence of localized plastic deformation and fracture-like features is consistent with hydrogen-assisted damage involving several interacting mechanisms. Hydrogen accumulation at grain boundaries and other microstructural trapping sites, including deformation twins, dislocation structures, and non-metallic inclusions, can promote localized plastic deformation through the HELP mechanism while simultaneously reducing local cohesive strength through HEDE, thereby facilitating crack initiation, branching, and propagation under cyclic contact loading [9,12,18,23]. More generally, hydrogen-assisted damage in other hydrogen-sensitive metallic systems has also been associated with hydrogen-enhanced nanoscale defect accumulation, nanopore formation, and subsequent crack initiation and propagation [66], although the specific microstructural mechanisms depend on the material system.

The crack-tip morphology observed in panels (g,h) shows a progressive narrowing of the crack and reduced branching toward its terminal region. These features indicate attenuation of crack development within the subsurface region; however, the present SEM observations do not allow the crack-arrest mechanism to be attributed directly to local hydrogen depletion or changes in the local stress state.

The observed crack morphology is consistent with the wear-intensity results presented in Figures 3 and 4. The extensive branching, local delamination, variable crack opening, and accompanying secondary microcracks observed in DDT68 after hydrogen charging and rolling–sliding testing indicate pronounced accumulated damage within the contact-affected near-surface region. Hydrogen retained at microstructural trapping sites can remain available for redistribution during subsequent mechanical loading and may therefore contribute to continued hydrogen-assisted deformation and crack development [59,62–64]. Previous investigations have also demonstrated that hydrogen can substantially modify the near-surface response of steels under machining and tribological conditions [66,67]. Moreover, the near-surface microstructural state can significantly affect wear resistance under hydrogen exposure [68], while surface nanostructuring has also been shown to influence hydrogen permeability in steels [69]. However, the present SEM observations do not permit the growth of the individual crack to be resolved quantitatively from one tribological cycle to another. Consequently, the observed crack morphology should be regarded as a qualitative indicator of accumulated hydrogen-assisted damage, consistent with the increased wear intensity measured for DDT68 after hydrogen charging.

The fine-grained austenitic microstructure of DDT68 steel (Figure 1), characterized by a large grain-boundary area and the presence of deformation twins, provides a high density of potential sites for hydrogen trapping and localized deformation [8,12,30,31]. Under rolling–sliding contact, hydrogen accumulation at such microstructural heterogeneities, combined with cyclic contact stresses, can promote localized plastic deformation and facilitate microcrack initiation and propagation [9,12,15,69,70]. The branched crack morphology and associated near-surface damage observed in DDT68 (Figure 5) are consistent with these hydrogen-assisted damage mechanisms and with the pronounced increase in wear intensity measured after hydrogen charging (Figure 4) compared with P900.

4.3. Mechanisms of Hydrogen-Assisted Rolling Contact Damage (HELP–RCF–Delamination Framework)

Under rolling contact conditions in a roller-on-roller configuration, repeated Hertzian loading generates high cyclic shear stresses within the near-surface and subsurface regions of the contacting bodies. Depending on the contact geometry, frictional conditions, and material response, these stresses can promote progressive accumulation of cyclic plastic deformation and microstructural damage beneath the contact surface [34,39]. With increasing numbers of loading cycles, localized subsurface damage may facilitate microcrack initiation, followed by crack propagation and coalescence and, ultimately, surface delamination, spalling, or flaking, representing characteristic manifestations of rolling contact fatigue (RCF) [34,39]. Recent microstructural investigations have further demonstrated that RCF damage can be accompanied by intense localized cyclic deformation and the development of white etching areas and white etching cracks under severe contact-loading conditions [34,35]. Thus, RCF damage evolution reflects the coupled effects of cyclic contact stresses, localized plastic deformation, microstructural transformation, and progressive crack development within the contact-affected region [34,35,39].

The presence of hydrogen can significantly modify the early stages of rolling-contact damage through hydrogen-enhanced localized plasticity (HELP). Within the HELP framework, hydrogen–dislocation interactions can enhance dislocation mobility and promote the localization of plastic deformation within highly stressed regions [9,71]. Under cyclic

rolling–sliding contact, such hydrogen-enhanced dislocation activity can intensify localized slip and increase deformation heterogeneity within the near-surface and subsurface regions subjected to high cyclic shear stresses. Consequently, rather than attributing crack initiation solely to the progressive accumulation of conventional cyclic plastic deformation, the presence of hydrogen should be considered as an additional factor promoting localized deformation and facilitating microcrack initiation at microstructural barriers and other highly stressed regions [9,71]. In this way, HELP can contribute to an earlier transition from distributed cyclic microplasticity toward localized damage accumulation and crack initiation, although the extent of this effect depends on the local hydrogen concentration, microstructure, and contact-stress state.

Although the HELP mechanism provides a useful framework for interpreting the present experimental observations, it is unlikely to operate in isolation. In cold-worked high-nitrogen austenitic steels, deformation twins, grain boundaries, non-metallic inclusions, and other microstructural heterogeneities can act as hydrogen-trapping sites and local stress concentrators, thereby creating conditions in which hydrogen-enhanced decohesion (HEDE) may also contribute to damage evolution [10,11,30,31]. Hydrogen accumulation at susceptible interfaces can reduce local cohesive strength and facilitate microcrack nucleation and propagation, particularly when combined with hydrogen-enhanced localized plastic deformation [10,11]. In addition, hydrogen–vacancy interactions and hydrogen-enhanced vacancy stabilization may contribute to vacancy clustering and nanovoid formation during plastic deformation, providing an additional pathway for damage nucleation [72–75]. Since the present study did not include atomistic characterization or high-resolution TEM analysis capable of directly distinguishing these mechanisms, HEDE and hydrogen–vacancy interactions are considered here as complementary mechanisms that may operate together with HELP rather than as mechanisms independently demonstrated by the present experiments.

Subsequently, localized cyclic shear deformation within highly stressed subsurface regions can promote the initiation of microcracks. Crack nucleation preferentially occurs at microstructural heterogeneities, including non-metallic inclusions, carbides/nitrides, grain boundaries, and deformation twins, many of which may also act as hydrogen-trapping sites and local stress concentrators [8,30,31]. Under such conditions, hydrogen accumulation can facilitate crack initiation and accelerate subsequent crack development through its interaction with localized plastic deformation and microstructural interfaces [8,10,11]. For bearing steels, experimental investigations have demonstrated that hydrogen uptake can substantially reduce rolling contact fatigue (RCF) life and promote microcrack formation, microstructural alteration, and the development of characteristic subsurface and near-surface damage that may ultimately contribute to flaking or spalling [6,33,43,49].

The proposed relationship between conventional rolling contact fatigue and hydrogen-assisted damage evolution through HELP-assisted deformation localization and subsequent subsurface crack development is schematically summarized in Figure 6.

At the RCF propagation stage, hydrogen can further accelerate damage evolution by promoting the formation, growth, and coalescence of microcracks within highly stressed subsurface and near-surface regions. Hydrogen-assisted localization of plastic deformation, together with hydrogen accumulation at microstructural traps and crack-tip regions, can facilitate crack propagation along locally weakened paths and promote the development of interconnected crack networks. Experimental studies on bearing steels have demonstrated that increased hydrogen content, originating either from lubricant degradation and tribochemical reactions during rolling contact or from prior hydrogen charging, can substantially reduce RCF life and accelerate the formation of white-etching crack systems, flaking, and spalling [6,33,43,49]. Consequently, hydrogen can shift the RCF damage pro-

cess toward earlier crack coalescence and premature material detachment compared with corresponding hydrogen-free conditions. The proposed hydrogen-assisted progression from microcrack formation and coalescence to interconnected subsurface crack networks and subsequent material spalling is schematically summarized in Figure 7.

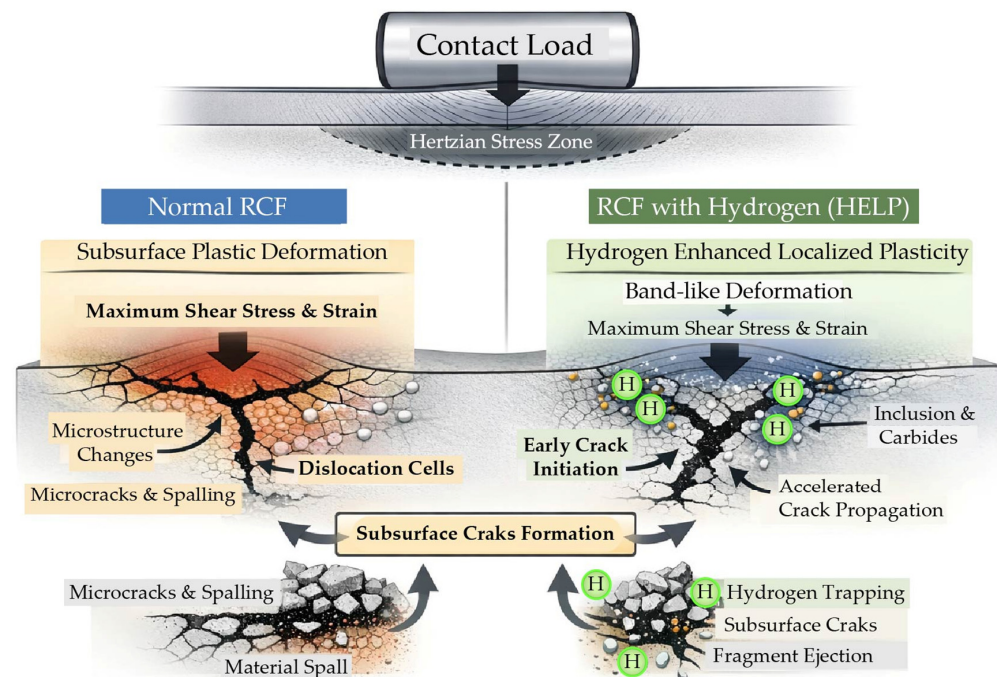


Figure 6. Mechanism of subsurface crack initiation and spalling in rolling contact fatigue: effect of hydrogen-enhanced localized plasticity (HELP).

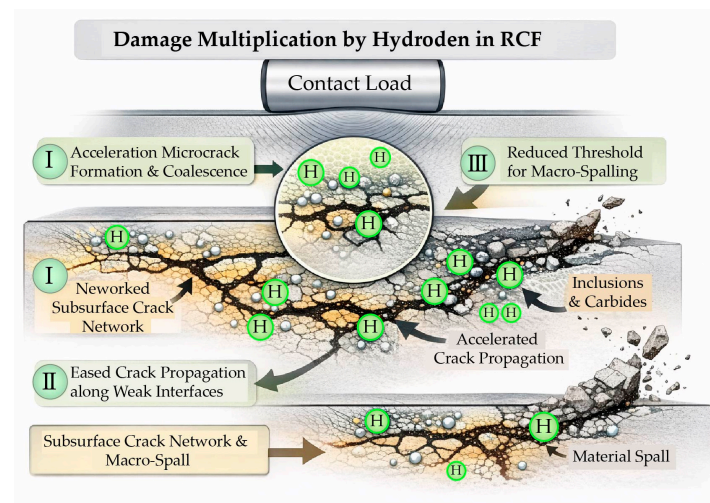


Figure 7. Hydrogen-enhanced formation and coalescence of subsurface crack networks leading to premature spalling under rolling contact.

The final stage of this degradation sequence involves delamination and the associated intensification of wear. According to the classical delamination theory of wear, repeated plastic deformation within the near-surface region can promote the nucleation and growth of subsurface cracks, including cracks propagating approximately parallel to the contact surface; their subsequent coalescence can ultimately result in the detachment of plate-like wear fragments [41]. In the presence of hydrogen, HELP-assisted localization of plastic deformation, together with hydrogen-assisted crack initiation and propagation, may accelerate the accumulation and coalescence of such subsurface damage [9,41,75].

Consequently, conditions favorable for delamination and material detachment may develop earlier, leading to increased microcrack density, enhanced formation of flake- or platelet-like wear debris, and ultimately an increased wear intensity. Within the present experiments, this mechanistic interpretation is consistent with the higher wear intensity and more pronounced crack development observed for hydrogen-charged DDT68 steel.

Thus, under the investigated rolling–sliding conditions, hydrogen-assisted damage in high-nitrogen austenitic steels can be interpreted as a sequence of mutually interacting processes: hydrogen-enhanced localization of plastic deformation through HELP → microcrack initiation within highly stressed near-surface and subsurface regions containing hydrogen traps and microstructural stress concentrators → crack growth and coalescence under cyclic rolling–contact loading → development of near-surface delamination and material detachment → increased wear intensity [49,72–75]. This sequence should be regarded as a mechanistic framework for interpreting the present experimental observations rather than as a directly resolved temporal sequence of damage events.

Within this framework, wear intensity, wear-debris characteristics, and crack-development parameters can be regarded not merely as consequences of tribological interaction, but as complementary indicators of hydrogen-assisted damage accumulation within the near-surface and subsurface regions of the material. This interpretation is consistent with studies demonstrating that hydrogen can accumulate locally in the vicinity of microstructural defects and stress concentrators, thereby increasing the susceptibility of these regions to damage initiation and development [76]. Consequently, the combined analysis of wear intensity, crack morphology, and wear-debris characteristics provides a useful basis for assessing the progression of hydrogen-assisted tribological degradation under the investigated rolling–sliding conditions.

This approach provides a physically motivated basis for moving from a qualitative mechanistic interpretation of hydrogen-assisted wear toward the formulation of quantitative wear-based degradation indicators. Previous studies have demonstrated that hydrogen concentration can be related to the residual durability of hydrogen-affected structural components [77], supporting the broader concept of incorporating hydrogen-related material degradation into quantitative durability assessments. In the present study, however, the proposed wear-based indicators are intended for comparative assessment of material degradation under the investigated rolling–sliding conditions rather than for direct prediction of component service life. The corresponding indicator-based framework is discussed in the following subsection.

4.4. Dimensionless Wear-Based Indicator for Comparative Assessment of Hydrogen-Assisted Degradation

To support a quantitative comparison of the susceptibility of the investigated steels to hydrogen-assisted degradation, a dimensionless indicator was formulated by combining the experimentally determined relative change in wear intensity after hydrogen charging with a normalized factor representing the measured hydrogen concentration. The indicator is intended exclusively for comparative assessment within the experimental conditions of the present study and should not be interpreted as a validated predictive model of component service life.

This approach is consistent with the classical concept that, to a first approximation, wear volume scales with the applied normal load and inversely with material hardness, as described by the Archard wear relation [78]. Under rolling–sliding conditions, however, deviations from a comparatively stable wear response may arise when cyclic contact damage, microcrack initiation and propagation, and material delamination increasingly contribute to material removal [34,35,39,41]. Accordingly, an increase in wear intensity can be considered as an experimentally measurable manifestation of the evolving wear–fatigue

damage process, although it should not be interpreted independently as a direct measure of remaining service life or proximity to failure.

At the same time, rolling-contact fatigue life is traditionally described using statistical and volumetric approaches, including the Lundberg–Palmgren model and its subsequent extension by Ioannides and Harris, in which fatigue-life assessment is related to the magnitude and distribution of subsurface stresses within the stressed or “risk” volume beneath the rolling contact [38,39]. These established RCF life models are referenced here only to provide a physical context for the strong dependence of rolling-contact damage on the subsurface stress state; the dimensionless indicator proposed in the present study is not intended to reproduce or replace probabilistic bearing-life calculations.

In this context, hydrogen represents an additional degradation factor that can substantially reduce RCF resistance even at relatively low concentrations in the material. Experimental investigations on bearing steels have demonstrated that hydrogen uptake can markedly shorten RCF life and accelerate the development of subsurface cracking, white-etching microstructural alterations, and premature flaking or spalling [6,33,43,49]. This experimentally established sensitivity of rolling-contact damage to hydrogen provides the physical rationale for incorporating a hydrogen-related factor into the comparative degradation indicator proposed in the present study.

Experimental measurements established hydrogen contents of 12–14 ppm for P900 and 15–17 ppm for DDT68. Thus, under identical hydrogen-charging conditions, DDT68 retained a higher hydrogen concentration than P900, indicating a greater hydrogen uptake under the conditions employed in the present study. Quantitative measurements of hydrogen in bearing steels subjected to tribo-mechanical loading have demonstrated that hydrogen uptake depends strongly on the loading and contact conditions [78], while experimental RCF studies have independently shown that increased hydrogen content can substantially reduce rolling-contact fatigue resistance and accelerate hydrogen-assisted damage development [6,33,43,49]. These observations provide a physical basis for including the measured hydrogen concentration as one component of the comparative degradation indicator used in the present study.

4.4.1. Definition of the Comparative Degradation Indicator

For the comparative assessment of hydrogen-assisted degradation, a dimensionless indicator D is defined as the product of a wear-related component, D_{wear} , and a hydrogen-related component, D_{H} :

$$D = D_{\text{wear}} \times D_{\text{H}} \quad (1)$$

where D_{wear} represents the wear-related component characterizing the change in wear response after hydrogen charging, and D_{H} represents the hydrogen-related component accounting for the measured hydrogen content.

The wear-related component is defined as:

$$D_{\text{wear}} = \left(\frac{I_{\text{H}}}{I_0} \right)^{\alpha}, \quad (2)$$

where I_0 is the wear intensity in the hydrogen-free (“baseline”) condition and I_{H} is the wear intensity after hydrogen charging. In the present study, the exponent was fixed at $\alpha = 1$, corresponding to the simplest linear normalization of the experimentally measured wear ratio. No independent parameter calibration was performed because the available experimental dataset did not allow separate calibration and validation. Therefore, the exponent should not be interpreted as a fitted material parameter. Such normalization directly reflects the experimentally observed change in wear intensity during hydrogen-assisted degradation and therefore provides a dimensionless measure of the hydrogen-associated wear response.

As subsurface crack formation and subsequent delamination increasingly contribute to material removal, an increase in wear intensity may be expected [41].

The hydrogen-related component is defined as:

$$D_H = \left(\frac{C_H}{C_{ref}} \right)^\beta, \quad (3)$$

where C_H is the experimentally measured hydrogen concentration (ppm), C_{ref} is the reference hydrogen concentration used for normalization, and β is a weighting exponent defining the contribution of the hydrogen-related component to the comparative indicator. In the present study, $C_{ref} = 10$ ppm was selected because it lies below the measured hydrogen concentration ranges obtained for both steels under identical electrochemical charging conditions and therefore provides a common baseline for their comparative normalization. Accordingly, C_{ref} should not be interpreted as a universal critical hydrogen concentration or a material-independent failure threshold. The exponent was fixed at $\beta = 1$, corresponding to the simplest linear normalization of the experimentally measured hydrogen concentration. No independent parameter calibration was performed because the available experimental dataset did not allow separate calibration and validation. Therefore, β should not be interpreted as a fitted material parameter, but rather as a simplifying assumption adopted within the present comparative indicator framework.

For the investigated steels, the experimentally measured hydrogen concentrations were:

$$C_H(\text{P 900}) = 12 - 14 \text{ ppm} \rightarrow D_H \approx (1.2 - 1.4)^\beta,$$

$$C_H(\text{DDT 68}) = 15 - 17 \text{ ppm} \rightarrow D_H \approx (1.5 - 1.7)^\beta,$$

Thus, under the adopted normalization, the hydrogen-related component D_H is higher for DDT68 than for P900, reflecting its greater experimentally measured hydrogen uptake. Accordingly, substitution of Equations (2) and (3) into Equation (1) gives the combined dimensionless degradation indicator:

$$D = \left(\frac{I_H}{I_0} \right)^\alpha \left(\frac{C_H}{C_{ref}} \right)^\beta. \quad (4)$$

For $\alpha = \beta = 1$, Equation (4) provides a simple dimensionless combination of the experimentally measured wear-intensity ratio and the normalized hydrogen concentration. Higher values of D therefore indicate a stronger hydrogen-associated degradation response within the investigated experimental conditions. The indicator is intended for comparative ranking of the investigated material states and does not represent a direct prediction of fatigue life or remaining service life.

Accordingly, the proposed indicator D integrates two experimentally determined aspects of hydrogen-assisted degradation—the change in wear intensity and the hydrogen uptake—into a single dimensionless quantity. Within the scope of the present study, it is used exclusively for comparative assessment of the relative hydrogen sensitivity of DDT68 and P900 steels under identical rolling-sliding and hydrogen-charging conditions. Its application beyond these experimental conditions would require independent calibration and validation.

4.4.2. Practical Implications and Limitations of the Indicator

From a practical perspective, the dimensionless indicator D may be used to compare the relative susceptibility of different material states to hydrogen-assisted tribological degradation under equivalent experimental conditions. Higher values of D indicate a

stronger combined contribution of hydrogen uptake and hydrogen-associated wear intensification. However, the present dataset does not provide a sufficient basis for establishing universal threshold values or discrete risk categories (e.g., low, medium, or high); such classification would require independent calibration and validation using a substantially larger experimental dataset.

Since hydrogen trapping and localized deformation in high-nitrogen steels are associated with microstructural heterogeneities such as grain boundaries, deformation twins, inclusions, and dislocation structures, the proposed indicator D may be complemented by morphological damage descriptors obtained from surface, subsurface, and wear-debris analyses. These may include the density and branching of subsurface cracks and the morphology and abundance of flake-like wear particles associated with delamination. Thus, combining the dimensionless degradation indicator with morphological information from crack and wear-debris analyses provides a potential basis for future condition-monitoring approaches.

To transition from the mechanistic interpretation of hydrogen-assisted damage (HELP $\rightarrow$ shear localization $\rightarrow$ subsurface crack initiation $\rightarrow$ rolling contact fatigue $\rightarrow$ delamination $\rightarrow$ wear) toward a quantitative comparison of the degradation response of high-nitrogen austenitic steels, an indicator-based comparative assessment framework is proposed. The framework combines the experimentally measured change in wear intensity after hydrogen charging with the measured hydrogen concentration in the material.

The tests were conducted using a roller–roller configuration. For the purpose of normalizing wear intensity to the sliding–rolling distance, one roller revolution corresponds to a circumferential travel distance L_{rev} , given by:

$$L_{rev} = \pi d, \quad (5)$$

where d is the roller diameter expressed in meters. Accordingly, the number of roller revolutions corresponding to a reference travel distance of 1000 m is:

$$n_{1000} = \frac{1000}{\pi d}, \quad (6)$$

For DDT68 ($d = 35$ mm), $n_{1000} = 9.09 \times 10^3$ revolutions per 1000 m, whereas for P900 ($d = 31$ mm), $n_{1000} = 1.03 \times 10^4$. These values are used solely to normalize the experimentally measured wear response with respect to the reference travel distance and should not be interpreted as fatigue-life cycles.

The indicator-based number of roller revolutions corresponding to a prescribed accumulated mass-loss reference level, Δm^* , was calculated according to the following relation:

$$N(\Delta m^*) = n_{1000} \times \frac{\Delta m^*}{I(P)}, \quad (7)$$

where $I(P)$ is the experimentally determined wear intensity at the applied contact load P , Δm^* is a prescribed accumulated mass-loss reference level, and n_{1000} is the number of roller revolutions corresponding to a reference travel distance of 1000 m, as defined by Equation (6).

For comparative engineering assessment, three prescribed accumulated mass-loss reference levels were considered: $\Delta m_I^* = 0.1$ g, $\Delta m_{II}^* = 0.3$ g, and $\Delta m_{III}^* = 0.5$ g. These values define three prescribed reference levels representing progressively increasing accumulated mass loss and are used solely for comparative engineering assessment. The corresponding quantities N_I , N_{II} , and N_{III} denote the indicator-based numbers of roller revolutions corresponding to these prescribed reference levels.

The selected reference levels were introduced exclusively as engineering reference levels to provide a consistent numerical basis for the comparative assessment of the investigated steels. They should not be interpreted as experimentally synchronized physical thresholds for the initiation of the first stable subsurface crack, the onset of delamination or spalling, or component failure. The identification of event-specific physical thresholds would require interrupted rolling-contact fatigue experiments combined with direct surface and subsurface observations performed at predefined test intervals. Because Equation (7) is linear with respect to the prescribed accumulated mass-loss level, the calculated reference-level counts satisfy $N_I:N_{II}:N_{III} = 1:3:5$ at each applied load for a given material. This proportionality reflects the mathematical structure of Equation (7) and was used as an internal consistency check for the calculated reference-level dependencies.

The reference-level revolution counts calculated according to Equation (7) for the prescribed accumulated mass-loss reference levels are summarized in Table 6.

Table 6. Calculated numbers of roller revolutions corresponding to the prescribed accumulated mass-loss reference levels according to Equation (7).

Load (N)	Material	Wear Intensity, $I(P)$ (g/1000 m)	N_I (0.1g)	N_{II} (0.3g)	N_{III} (0.5g)
200	DDT68	0.063	14,436	43,307	72,179
300	DDT68	0.175	5197	15,591	25,984
400	DDT68	0.527	1726	5178	8663
500	DDT68	1.001	909	2727	4547
600	DDT68	1.753	520	1556	2593
200	P900	0.035	29,337	88,011	146,684
300	P900	0.112	9168	27,505	45,842
400	P900	0.444	2313	6943	11,566
500	P900	0.725	1416	4249	7082
600	P900	1.523	674	2027	3371

After electrochemical hydrogen charging, the measured hydrogen concentration was 12–14 ppm for P900 steel and 15–17 ppm for DDT68 steel. At applied loads between 400 and 600 N, the experimentally determined wear intensity increased markedly after hydrogen charging, indicating a stronger hydrogen-associated wear response under higher contact loads. This behavior is consistent with the observed crack morphology and delamination features and with the mechanistic interpretation of hydrogen-assisted rolling-contact damage discussed above. Consequently, the calculated indicator-based dependencies $N_I(P)$, $N_{II}(P)$, and $N_{III}(P)$, presented in Figure 8, provide a comparative quantitative representation of the roller-revolution counts corresponding to the prescribed accumulated mass-loss reference levels rather than predictions of discrete physical damage events or component service life.

Thus, the experimentally measured wear intensity, considered together with the measured hydrogen content, provides a quantitative basis for the comparative assessment of hydrogen-assisted degradation susceptibility under the investigated rolling–sliding contact conditions. Within the proposed indicator framework, these parameters enable comparative ranking of the investigated material states without implying direct prediction of fatigue life, remaining service life, or specific physical damage thresholds.

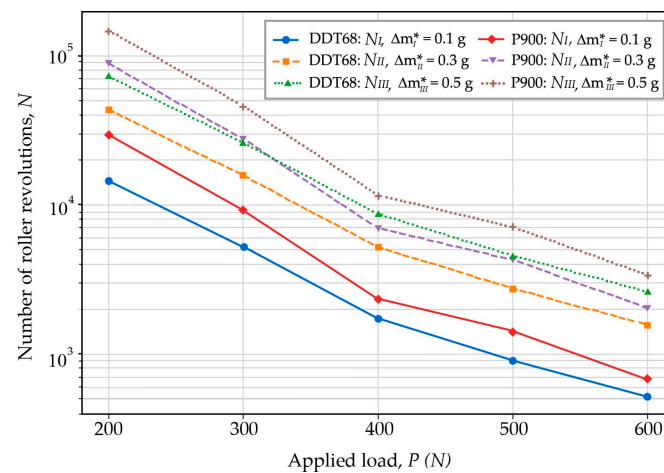


Figure 8. Calculated roller-revolution dependencies on applied load for DDT68 and P900 steels corresponding to the prescribed accumulated mass-loss reference levels $\Delta m_I^* = 0.1$ g, $\Delta m_{II}^* = 0.3$ g, and $\Delta m_{III}^* = 0.5$ g under dry rolling–sliding contact after hydrogen charging. The prescribed accumulated mass-loss values are engineering reference levels used exclusively for comparative assessment and do not represent physical thresholds for crack initiation, delamination, spalling, or component failure. **The asterisk in Δm is part of the parameter notation and denotes the prescribed accumulated mass-loss reference level; it does not indicate statistical significance.**

4.5. Strategies for Mitigating Hydrogen-Assisted Wear and Rolling Contact Fatigue in High-Nitrogen Austenitic Steels

Hydrogen-assisted wear and rolling contact fatigue (RCF) in high-nitrogen austenitic steels involve a complex interaction among contact stresses, cyclic plastic deformation, microstructural heterogeneities, and hydrogen transport and trapping. Under rolling–sliding contact, repeated Hertzian loading generates highly stressed near-surface and subsurface regions in which localized plastic deformation, microstructural damage, microcrack initiation and coalescence, and ultimately spalling may occur [39,40]. The presence of hydrogen can further modify this damage evolution through its interaction with dislocations and microstructural trapping sites, promoting localized deformation and facilitating crack development [8–10,30,31]. Experimental investigations on bearing steels have demonstrated that hydrogen can accelerate subsurface microcrack development, microstructural alteration, and rolling-contact damage and can substantially reduce RCF resistance [3–6,33,43,49].

Hydrogen has also been shown to accelerate microstructural alterations associated with rolling contact fatigue, including the formation of white etching areas (WEAs) and white etching cracks (WECs) in bearing steels. Hydrogen–dislocation interactions can enhance localized deformation and dislocation-assisted carbon redistribution, thereby promoting the development of hydrogen-influenced microstructural alterations and associated crack networks under severe rolling-contact conditions [34,35,43,49]. For high-nitrogen austenitic steels, nitrogen alloying can reduce hydrogen diffusivity and decrease the depth of the hydrogen-affected near-surface region, thereby modifying hydrogen transport and redistribution within the material [18]. Nevertheless, the susceptibility of Cr–Mn–N austenitic steels to hydrogen-assisted degradation remains dependent on hydrogen exposure conditions, austenite stability, and the operative deformation and slip mechanisms [14,18]. At the nanoscale, this interpretation is consistent with mechanistic models in which hydrogen accumulation in highly stressed regions near crack tips interacts with dislocation processes and can reduce the ability of the material to accommodate crack-tip deformation, thereby facilitating crack propagation [15].

Accordingly, mitigation of hydrogen-assisted wear and RCF in high-nitrogen austenitic steels should combine microstructural, tribological, and diagnostic approaches. From a

materials-design perspective, control of austenite stability, grain structure, deformation substructure, and the population of hydrogen-trapping sites may reduce excessive hydrogen accumulation and suppress strongly localized plastic deformation [14,18]. From a tribological perspective, optimization of contact conditions, surface integrity, frictional interactions, and lubrication strategies can reduce the severity of near-surface damage and modify the wear response of hydrogen-exposed high-nitrogen steels [16]. Previous studies on nitrogen–manganese steels under rolling friction have demonstrated that modification of the contact zone can substantially affect their tribotechnical response [78]. Finally, monitoring the morphology and evolution of wear debris provides a complementary route for detecting changes in the degradation process; computer-vision-based analysis of wear products may facilitate the development of automated condition-monitoring approaches for hydrogen-exposed tribological systems [79,80]. These combined strategies provide a practical framework for reducing hydrogen-assisted damage while simultaneously improving the detectability of its early tribological manifestations.

4.5.1. Microstructural Design: Stabilizing Austenite, Controlling Planar Slip and Trap Hierarchy

In high-nitrogen austenitic steels of the Cr–Mn–N system, susceptibility to hydrogen-assisted degradation is strongly influenced by austenite stability, the operative deformation mode, and the distribution of hydrogen-trapping sites. Experimental studies on stable austenitic steels have shown that deformation dominated by planar slip or mechanical twinning can be associated with greater hydrogen-induced ductility loss than deformation characterized by a higher contribution of dislocation cross-slip [14,40]. Dislocations, deformation twins, grain boundaries, precipitates, and non-metallic inclusions can additionally act as hydrogen-trapping sites and promote heterogeneous hydrogen redistribution within the microstructure [8,18,30,31]. Nitrogen alloying can decrease hydrogen diffusivity and reduce the depth of the hydrogen-affected region, while the resulting changes in deformation behavior and hydrogen transport can influence the localization of hydrogen-assisted damage [18]. Accordingly, microstructural design strategies should focus on maintaining sufficient austenite stability, controlling grain size and deformation substructure, and tailoring the type and spatial distribution of hydrogen-trapping sites to limit strongly localized hydrogen accumulation and plastic deformation [14,18,40].

This microstructural-design framework is supported by experimental evidence on hydrogen environment embrittlement (HEE) in Cr–Mn–N and other austenitic steels, demonstrating that hydrogen susceptibility is strongly influenced by alloy composition, austenite stability, temperature, and the operative deformation mechanisms [14,40]. In particular, the deformation mode, including the relative contributions of planar slip, mechanical twinning, and dislocation cross-slip, can substantially influence hydrogen-induced ductility loss and fracture behavior in stable austenitic steels [40]. Nitrogen alloying provides an additional microstructural control parameter because it can decrease hydrogen diffusivity and modify the depth of the hydrogen-affected region and the localization of deformation [18]. Furthermore, studies on metastable austenitic stainless steels demonstrate that control of phase stability and deformation-induced transformation behavior can improve mechanical strength without necessarily compromising hydrogen embrittlement resistance [42]. Collectively, these findings emphasize that austenite stability, deformation mode, hydrogen transport, and local microstructural heterogeneity should be considered jointly when designing hydrogen-resistant high-nitrogen austenitic steels [14,18,40,42].

In the context of the present DDT68–P900 comparison, these considerations indicate that hydrogen-assisted tribological degradation cannot be attributed to grain size alone, but should be interpreted as the combined result of grain structure, deformation substructure,

hydrogen-trapping capacity, and the resulting localization of plastic deformation under rolling–sliding contact.

4.5.2. Surface Severe Plastic Deformation and Residual Compressive Stresses: Shot Peening, Ultrasonic Peening, Laser Shock Peening

In the subsurface region of components subjected to rolling contact loading, the early stages of hydrogen-assisted wear and rolling contact fatigue (RCF) are initiated within the near-surface volume, where cyclic contact stresses and hydrogen locally accumulate. One of the most effective strategies to mitigate such damage is the formation of a compressive residual stress layer combined with surface strengthening. These modifications can reduce the effective tensile driving forces for crack initiation and opening and may alter hydrogen transport and trapping within the mechanically modified near-surface region. Previous studies have demonstrated that surface stresses and nanostructural modification can influence hydrogen absorption, diffusion, permeability, and hydrogen-embrittlement resistance [62,68,69]. Shot peening has also been shown to generate compressive residual stresses in stainless steel surface layers [81], while ultrasonic shot peening can modify the near-surface stress and microstructural state in a manner beneficial to resistance against environmental hydrogen embrittlement [82].

For the investigated roller pairs, the calculated maximum Hertzian subsurface shear stresses are located at depths of approximately 0.035–0.062 mm below the contact surface over the applied load range of 200–600 N. This depth interval should therefore be considered when selecting mechanical surface-strengthening parameters. Published experimental studies have shown that shot peening can generate pronounced compressive residual stresses within the near-surface region, with the maximum compressive stresses typically occurring within the first few hundred micrometers and the effective compressive layer extending deeper depending on the material and peening conditions [81,82]. Deep rolling can generally produce a deeper compressive residual-stress field, extending to several hundred micrometers or more under sufficiently high rolling forces [83,84]. Thus, appropriately selected shot-peening or deep-rolling parameters may generate a compressive residual-stress field that overlaps the calculated Hertzian shear-stress maximum in the present study, with deep rolling potentially providing a greater depth margin. From an engineering perspective, the treatment should therefore be optimized to ensure that the compressive residual-stress field encompasses the calculated damage-prone region (0.035–0.062 mm), thereby reducing the local tensile driving force for crack opening and potentially delaying the initiation and propagation of hydrogen-assisted subsurface cracks. Nevertheless, the magnitude, penetration depth, and cyclic stability of the induced residual stresses should be experimentally verified for the investigated Cr–Mn–N steels, since these characteristics depend strongly on the treatment parameters and material microstructure.

It has been emphasized that restricting surface strengthening to a very thin layer may be insufficient, since the shear-stress maxima relevant to rolling contact fatigue are located in the subsurface region, where an adequately deep compressive residual-stress field is desirable [83,84]. Moreover, post-treatment surface roughness should be carefully controlled, because mechanical surface treatments can modify both the residual-stress state and surface topography [44,81]. Experimental studies on shot peening and deep rolling indicate that the resulting balance between fatigue performance, residual-stress development, and surface-condition modification depends strongly on the selected processing parameters and material response [44,81,83,84].

For the practical implementation of these approaches in high-nitrogen steels (HNS) under rolling contact conditions, the following measures are recommended:

- to ensure the formation of a sufficiently deep compressive residual-stress layer rather than merely superficial hardening, so that the strengthened region encompasses the subsurface Hertzian shear-stress maximum associated with RCF damage initiation;
- to control post-treatment surface roughness, since excessive roughness may adversely modify the local contact conditions and partially offset the beneficial effect of the induced compressive residual stresses;
- to verify the resulting material state by residual-stress depth profiling using X-ray diffraction (XRD) and depth-resolved microhardness measurements in order to assess the magnitude, penetration depth, and stability of the strengthening effect.

Mechanical surface modification can additionally alter the density and distribution of dislocations, grain boundaries, and other microstructural defects that act as hydrogen-trapping sites [68,69]. However, the resulting effect on hydrogen-assisted degradation is not necessarily unconditionally beneficial: depending on trap binding energy, density, distribution, and reversibility, such defects may either reduce hydrogen mobility or promote local hydrogen accumulation in highly stressed regions [8,30,70]. Therefore, optimization of surface strengthening for HNS should consider not only the depth and magnitude of compressive residual stresses but also the associated changes in hydrogen transport and trapping behavior. Mechanical pulse treatment provides an additional example of a surface-modification route capable of improving the corrosion and wear resistance of structural steels [85].

Overall, the present results suggest that the effectiveness of mechanical surface strengthening for hydrogen-charged HNS should be assessed not simply by the increase in surface hardness, but by the correspondence between the depth of the modified layer and the subsurface region subjected to the highest cyclic contact stresses. For the DDT68 and P900 roller pairs investigated here, the calculated Hertzian shear-stress maxima at approximately 0.035–0.062 mm provide a physically meaningful target depth for the compressive residual-stress field. Accordingly, shot peening, ultrasonic peening, laser shock peening, or deep rolling may be considered promising mitigation strategies provided that the induced compressive stress field extends through this damage-prone region without introducing excessive surface roughness or unfavorable hydrogen-trapping conditions. Experimental validation of these treatments on DDT68 and P900 under hydrogen-assisted rolling–sliding contact therefore represents an important direction for further work.

4.5.3. Generalized Strategies for Enhancing the Service Life of HNS Under Hydrogen-Containing Rolling Contact Conditions: From Surface Modification to Diagnostics

Hydrogen-assisted wear and rolling contact fatigue (RCF) in high-nitrogen austenitic steels (HNS) develop within a mechanically and chemically heterogeneous near-surface/subsurface region, where cyclic contact stresses, defect-substructure evolution, and hydrogen transport and trapping interact. Therefore, an integrated (“duplex/multi-layer”) strategy can be considered a promising approach aimed at simultaneously: (I) reducing the effective tensile driving forces for crack initiation and opening; (II) limiting hydrogen ingress and detrimental local accumulation; (III) stabilizing the tribological regime and limiting conditions that may promote additional hydrogen uptake; and (IV) enabling early detection of the transition from distributed microcracking to macroscopic delamination or spalling.

(I) Low-Temperature Thermochemical Treatments (Expanded Austenite, S-Phase).

For austenitic steels, low-temperature nitriding or carburizing leading to the formation of expanded austenite (S-phase, $\gamma\text{N}/\gamma\text{C}$) represents a promising surface-engineering approach. Such treatments can substantially increase surface hardness while retaining good corrosion resistance when chromium-nitride precipitation is suppressed [86–88]. The

resulting hardened layer can reduce near-surface plastic deformation and improve wear resistance, while the high interstitial content and associated lattice expansion generate a modified near-surface stress and defect state [86–88]. From the perspective of hydrogen-assisted degradation, however, the influence of expanded austenite on hydrogen ingress, transport, and trapping should be considered separately, since these processes depend on the composition, defect structure, trap characteristics, and stability of the modified layer [8,30,70]. Therefore, the application of expanded-austenite treatments to HNS under hydrogen-containing rolling-contact conditions requires direct experimental verification rather than extrapolation solely from conventional wear and corrosion studies.

(II) limiting hydrogen ingress and detrimental local accumulation.

In parallel with surface strengthening, hydrogen-barrier coatings with low hydrogen permeability may provide an additional means of limiting hydrogen ingress. Their effectiveness, however, depends not only on the intrinsic permeability of the coating material but also on coating continuity, defect density, adhesion, and mechanical stability, since pores, cracks, and local coating damage can provide preferential pathways for hydrogen transport [87–89].

From a practical tribological perspective, duplex surface-engineering strategies, such as an expanded-austenite or nitrided supporting layer combined with a hard PVD or DLC-based topcoat, may offer advantages over a single thin hard coating [90,91]. The hardened subsurface layer can improve load-bearing capacity and provide mechanical support for the topcoat under repeated loading [90,91], whereas the topcoat may contribute to reducing friction and wear [90–93]. When specifically designed as a hydrogen-permeation barrier, a sufficiently continuous and defect-controlled coating may additionally limit hydrogen ingress [87–89]. However, the overall effectiveness of such architectures depends strongly on interfacial adhesion, coating integrity, residual stresses, and mechanical compatibility between the modified substrate and the topcoat [90,91,93].

As an industrially relevant barrier option, electrolytic and electroless Ni–P coatings have attracted particular attention because of their potential to reduce hydrogen permeation. A recent dedicated review summarizes the mechanisms, processing routes, microstructural factors, and barrier performance of Ni and Ni–P coatings and emphasizes the importance of coating continuity and defect control for achieving effective hydrogen-permeation resistance [92].

For low-friction protection and friction stabilization, DLC-based approaches are also systematically considered, with particular emphasis on coating adhesion and tribo-oxidation control [93].

For HNS under rolling–sliding contact, however, reducing hydrogen ingress alone may not be sufficient, because hydrogen already present in the material can redistribute among lattice sites and microstructural traps under cyclic deformation and locally elevated stresses [8,30,70]. Therefore, an effective barrier or duplex strategy should combine reduced hydrogen entry with control of the subsurface microstructural state and preservation of coating integrity under repeated contact loading. In the context of the present results, particular attention should be paid to preventing hydrogen accumulation within the near-surface/subsurface region where the maximum Hertzian shear stresses and hydrogen-assisted damage processes overlap.

(III) Tribochemistry and Lubrication Management (Control of “Contact” Hydrogen).

For low-friction surface protection and stabilization of the tribological regime, DLC-based coatings represent a promising approach, particularly when coating adhesion, wear resistance, and tribo-oxidation stability are adequately controlled [93]. However, under hydrogen-containing rolling-contact conditions, friction reduction and surface protection alone may be insufficient, because premature damage is also governed by the evolution of

the subsurface stress state and hydrogen trapping. Recent investigations of bearing steels have demonstrated a close relationship between residual-stress evolution, hydrogen trapping, and premature rolling-contact damage [94]. Furthermore, studies of rolling contact fatigue and white etching cracks (WECs) emphasize that hydrogen-related processes can interact with cyclic subsurface stresses and microstructural evolution, thereby contributing to the development of subsurface damage and premature bearing failure [95]. These findings indicate that friction control, surface protection, residual-stress management, and hydrogen transport and trapping should be considered as interconnected elements of an integrated strategy for mitigating hydrogen-assisted rolling-contact degradation.

From an engineering perspective, the mechanical and physicochemical stability of the friction interface is equally important. Experimental investigations of electric-spark coatings under boundary-lubrication conditions have demonstrated that surface-engineered layers and the evolution of secondary surface structures can substantially influence friction and wear behavior [96]. Studies of electrodeposited Ni–Mo–B coatings further demonstrate the close relationship between coating characteristics, wear resistance, corrosion resistance, and tribocorrosion performance [97]. Earlier investigations of austenitic manganese steels and cast irons under sliding-friction conditions likewise showed that their triboengineering behavior is strongly dependent on material characteristics and friction conditions [98]. Although these material systems and loading configurations differ from the HNS rolling–sliding contact considered in the present study, collectively these findings emphasize that surface state, coating integrity, and interfacial contact conditions should be controlled as interconnected rather than independent factors.

The importance of controlling the contact-zone conditions becomes particularly pronounced in hydrogen-containing environments. Investigations of friction components operating under hydrogen-containing conditions have demonstrated that hydrogen exposure can substantially affect wear resistance and the evolution of friction-induced surface degradation [99]. Studies of nitrogen steels exposed to hydrogen-containing fuels and operated with environmentally friendly lubricating coolants further indicate that the combined effects of material characteristics, hydrogen-containing environments, lubrication conditions, and frictional interactions can significantly influence tribological performance [100]. Although the investigated components and operating conditions differ from the HNS rolling–sliding system considered in the present study, these findings support the view that hydrogen should be regarded not only as a factor affecting bulk material degradation but also as an active variable influencing the evolution of the tribological interface.

A further critical mechanism is the generation of hydrogen directly within the lubricated contact. Rolling–sliding experiments have demonstrated that lubricant decomposition at mechanically activated metallic surfaces can generate hydrogen and that this process depends on lubricant chemistry and contact conditions [101]. Accordingly, lubricant formulation can influence the susceptibility of rolling bearings to premature damage associated with white etching cracks [102], while studies of AISI 52100 bearing steel have shown that lubricant characteristics also affect the development of white etching areas under cyclic loading [103]. Furthermore, investigations using different synthetic oils under rolling/sliding conditions have demonstrated that base-oil chemistry can influence hydrogen permeation into the contacting steel [104]. Consequently, lubricant selection for hydrogen-sensitive rolling contacts should consider not only viscosity, friction, and conventional antiwear performance, but also the propensity of the lubricant–surface system to generate, promote, or suppress tribochemically produced hydrogen.

(IV) Diagnostics and Condition Monitoring: Wear Debris Analysis and Indicator-Based Life Models.

Since hydrogen-assisted wear and RCF may progress through the development of distributed microcracking and wear-particle generation before macroscopic surface failure, wear-debris monitoring provides a potentially useful complementary approach for detecting changes in the degradation process. Recent reviews of on-line oil wear-debris detection technologies demonstrate that particle concentration, size, morphology, and related characteristics can provide diagnostically useful information on the evolving condition of tribological components [105].

For online condition-monitoring architectures, impedance-based microsensors provide a promising approach for detecting metallic wear debris in lubricating media. Experimental studies have demonstrated the feasibility of impedance-based microsensors for monitoring metallic debris in hydraulic oil, supporting their potential integration into automated wear-monitoring systems [106].

The diagnostic value of wear debris is particularly relevant to high-nitrogen manganese steels. Earlier investigations demonstrated that hydrogenation can substantially modify both the wear rate and the size and morphology of wear particles generated during friction, indicating that debris characteristics reflect changes in the underlying degradation mechanism [107]. These observations provide a direct experimental basis for considering wear-particle characteristics as complementary indicators of hydrogen-assisted tribological degradation in HNS.

Beyond conventional particle counting, image-based characterization and machine-learning methods provide additional opportunities for automated identification of evolving wear mechanisms. Deep-learning approaches have been successfully applied to ferrogaphic images for automatic segmentation and classification of wear particles [108], while knowledge-guided convolutional neural networks have demonstrated the ability to distinguish morphologically similar fatigue and severe-sliding wear debris using three-dimensional particle-surface information [109]. More recent online imaging architectures combining wear-particle sensing with adaptive deep learning further demonstrate the feasibility of real-time condition monitoring in rotating machinery [110]. These developments are consistent with earlier computer-vision approaches proposed for automated identification of wear products using image processing, artificial intelligence, and data-analysis methods [79,80].

Image-based diagnostics may also extend beyond detached debris to direct characterization of damage features on material surfaces. Computer analysis of fractographic images has been used to identify and quantify characteristic fracture features [111], while quantitative image analysis of material surfaces has been applied to evaluate crack characteristics [112]. Integration of such approaches with wear-debris sensing may therefore provide complementary information on both particle morphology and the evolution of surface and subsurface cracking during hydrogen-assisted degradation.

Within the framework adopted in the present study, such diagnostic information may be combined with the proposed indicator-based cycle metrics to provide a comparative, rather than predictive, assessment of degradation severity. The reference levels N_I , N_{II} , and N_{III} should therefore be interpreted as engineering indicators associated with prescribed normalized wear levels, rather than as independently validated physical lives to crack initiation or spalling. In a future condition-monitoring architecture, changes in wear rate, particle concentration, particle size and morphology, image-based crack descriptors, and other debris characteristics could be evaluated together to identify transitions in the degradation regime. However, calibration of such a framework against independently measured crack initiation, subsurface damage, and spalling events would be required before it could be used for quantitative remaining-life prediction.

(V) Link to RCF Durability Frameworks: From Delamination to Hydrogen-Assisted Damage Accumulation.

From a mechanistic perspective, the delamination theory of wear provides a useful framework for interpreting the transition from subsurface damage accumulation to material separation under repeated contact loading. Cyclic Hertzian stresses promote subsurface plastic deformation and crack nucleation, after which cracks may propagate approximately parallel to the surface and coalesce, ultimately producing platelet-like debris, delamination, and spalling [40,41]. These mechanisms are consistent with established descriptions of rolling contact fatigue, in which the evolution of subsurface stress fields and accumulated damage governs the progressive development of contact-induced failure [39,40]. Recent rolling–sliding studies further demonstrate that flake-type wear debris can provide direct evidence of progressive material separation associated with rolling-contact damage [14].

Hydrogen can accelerate this conventional RCF sequence by promoting localized plastic deformation, facilitating microcrack initiation, and enhancing the development of subsurface damage. Experimental studies on bearing steels have demonstrated substantial reductions in RCF resistance in hydrogen-containing environments and have directly linked hydrogen to accelerated microcrack formation and rolling-contact damage [3–6]. Hydrogen has also been associated with microstructural changes and surface-originated flaking under rolling contact [33], while recent studies of WEC/WEA evolution highlight the roles of subsurface shear localization and microstructural transformation in premature bearing damage [34,35]. Particularly relevant to the present HNS context, white etching areas have recently been identified in the high-nitrogen bearing steel X30CrMoN15-1, indicating that such RCF-related microstructural alterations are not restricted to conventional bearing-steel grades [36].

The hydrogen contribution to durability should therefore be considered in terms of both the applied cyclic stress field and the evolving local hydrogen–defect state. Experimental evidence shows that hydrogen charging and loading conditions can substantially alter the tribotechnical response of steels [58], while hydrogen-assisted nanoscale damage may progress from nanopore formation to crack initiation and propagation [75]. Moreover, local hydrogen enrichment in the vicinity of structural defects can differ substantially from the nominal bulk hydrogen content [76], and the residual durability of defect-containing steel components has been shown to depend on hydrogen concentration in the metal [77]. These findings support the use of hydrogen concentration, trapping state, and defect-assisted localization as additional degradation variables when conventional RCF durability concepts are extended to hydrogen-containing service conditions.

Classical RCF and bearing-life frameworks remain an appropriate mechanical backbone for such durability assessment because they relate cyclic contact stresses and accumulated damage to fatigue life [39,45]. However, these frameworks do not explicitly incorporate hydrogen concentration, hydrogen trapping, local hydrogen enrichment, or hydrogen-assisted microstructural evolution. Accordingly, the indicator-cycle relationships proposed in the present study should not be interpreted as calibrated predictions of the physical number of cycles to crack initiation or spalling. Rather, they provide a comparative engineering framework in which prescribed normalized wear levels are used to compare the relative degradation response of the investigated HNS. Quantitative prediction of absolute RCF life under hydrogen exposure would require independent calibration against experimentally determined crack-initiation, subsurface-damage, and spalling events, together with explicit consideration of contact stress, hydrogen concentration and trapping, residual stresses, and microstructural evolution.

4.6. Future Research Directions and Perspectives for High-Nitrogen Austenitic Steels in Hydrogen Energy Systems

Future development of high-nitrogen austenitic steels for hydrogen-energy tribological applications should focus on the simultaneous optimization of chemical composition, microstructure, hydrogen transport, and contact-damage resistance. Grain refinement cannot be regarded as universally beneficial under hydrogen exposure, because its strengthening effect may be accompanied by an increased density of hydrogen-trapping sites and stronger deformation localization [24–29]. Future alloy-design strategies should therefore establish an optimum combination of N, Cr, Mn, Ni, and Mo contents with grain size, austenite stability, deformation-twin and dislocation structures, secondary phases, and hydrogen-trap populations [20,21,59,63–65]. Particular attention should also be paid to carbides and non-metallic inclusions, since their amount, morphology, and distribution can influence wear, fatigue endurance, and microcrack initiation in highly loaded steels [112,113].

Reliable qualification of HNS for hydrogen service additionally requires careful control of hydrogen concentration and pressure, temperature, loading rate, exposure history, and testing procedure. Methodological investigations have demonstrated that the measured hydrogen resistance of steels depends strongly on testing conditions and that cyclic characteristics may be particularly sensitive to hydrogen-assisted degradation [114]. Studies of austenitic and martensitic steels in hydrogen further demonstrate pronounced temperature-dependent changes in mechanical properties and durability [115]. Future HNS investigations should therefore extend the present room-temperature electrochemical-charging approach toward gaseous hydrogen, elevated temperatures, different hydrogen pressures and concentrations, and cyclic rolling-contact loading representative of actual hydrogen-energy equipment.

Surface engineering represents another major research direction. Surface severe plastic deformation, low-temperature nitriding, expanded-austenite treatments, duplex architectures, and hydrogen-barrier coatings can modify surface hardness, residual stresses, load-bearing capacity, wear resistance, and hydrogen ingress [68,69,81–93,96,97,116]. Future studies should assess these characteristics simultaneously with subsurface crack evolution, since improvement in conventional wear resistance does not necessarily imply improved resistance to hydrogen-assisted RCF. Lubrication should likewise be considered as part of an integrated material–surface–environment system, because lubricant chemistry and tribochemical degradation can influence hydrogen generation, permeation, and white-etching damage under rolling–sliding contact [101–104].

Advanced manufacturing provides an additional route for tailoring HNS microstructures. Thermomechanical processing can be optimized through processing-map approaches to control deformation and microstructural evolution in austenitic steels [117], whereas additive manufacturing introduces additional variables associated with porosity, residual stresses, heterogeneous substructures, and high defect densities. Hydrogen-dependent deformation behavior has already been demonstrated for additively manufactured austenitic steels and Ni–Co superalloys [23,61]. Future work should therefore establish explicit processing–microstructure–hydrogen–tribology relationships and compare conventionally and additively manufactured HNS under identical hydrogen exposure and rolling–sliding conditions.

A further priority is the development of multiscale diagnostics linking local hydrogen accumulation and structural degradation to measurable condition indicators. Hydrogen-induced degradation can evolve across macro-, micro-, and nanoscales [2]. For tribological components, wear-debris concentration, particle size and morphology [42,105–107], impedance-based sensing [105,106], computer-vision analysis [79,80], machine-learning classification [108,109], and quantitative fracture- and crack-image analysis [110,111] pro-

vide complementary diagnostic information. Future monitoring systems should integrate these descriptors with hydrogen content, residual stresses, vibration and acoustic signals rather than relying on a single diagnostic parameter.

This requirement is particularly relevant to real hydrogen-energy machinery. Experience with hydrogen-cooled turbogenerators has demonstrated the practical significance of hydrogen-related degradation in rotating energy equipment [118], while investigations of rotor steels emphasize crack resistance and fracture behavior as critical considerations in selecting materials for hydrogen-targeted applications [119]. Recent engineering studies of compressor systems for hydrogen-cooled turbogenerators further demonstrate the continuing technological relevance of hydrogen-containing working environments in high-power electrical machinery [120]. More generally, durability studies of highly loaded turbine components demonstrate the importance of integrating stress, temperature, vibration, and fatigue-related parameters into service-life assessment [121]. These developments support a transition from isolated laboratory measurements toward multivariable structural-integrity assessment of HNS components under realistic hydrogen-energy operating conditions.

Intelligent processing of such heterogeneous data provides a natural pathway toward Industry 4.0/5.0 condition-monitoring and digital-twin architectures. Computer-vision and machine-learning methods can support automated recognition of wear particles, surface morphology, cracks, and changes in tribological behavior [79,80,108–111]. Recent data-driven tribological studies further demonstrate that machine-learning algorithms can identify nonlinear relationships between experimental contact parameters and frictional response [122]. Indicator-based approaches developed in automated engineering systems illustrate the broader methodological principle of transforming measurable material and process characteristics into quantitative decision-support parameters [123]. More broadly, information-support approaches developed for energy-oriented systems demonstrate the importance of integrating heterogeneous data into structured decision-support frameworks [124]. Although these approaches originate from applications different from hydrogen tribology, their underlying concepts of sensor-derived indicators, automated data processing, information integration, and decision support are transferable to future HNS monitoring systems.

Ultimately, wear intensity, debris morphology, hydrogen content, residual stresses, crack descriptors, friction parameters, and vibration or acoustic signatures could be combined within a continuously updated digital representation of a tribological component. Such an architecture would enable experimental observations at different scales to be integrated with material-state and operating-condition data. However, the indicator methodology proposed in the present study should initially remain a comparative engineering tool rather than being interpreted as a validated remaining-life predictor.

5. Conclusions

1. Hydrogen significantly modifies the rolling–sliding wear behavior of cold-worked high-nitrogen Cr–Mn–N austenitic steels by promoting localized plastic deformation and accelerating subsurface microcrack initiation within the Hertzian stress field, followed by crack coalescence, rolling contact fatigue, spalling, and delamination wear. The observed behavior is consistent with hydrogen-enhanced localized plasticity (HELP), with possible complementary contributions from hydrogen-enhanced decohesion (HEDE) and hydrogen–defect interactions. These results demonstrate the important role of hydrogen in subsurface damage evolution under dry rolling–sliding contact conditions.
2. The investigated steels exhibited markedly different hydrogen uptake depending on their microstructural state. The fine-grained DDT68 steel, with an average grain size of

approximately 25 μm , accumulated 15–17 ppm hydrogen, whereas the coarser-grained P900 steel, with an average grain size of approximately 43 μm , accumulated 12–14 ppm under identical electrochemical charging conditions. These results are consistent with a greater density of potential hydrogen-trapping sites in the finer-grained cold-worked microstructure and the correspondingly higher hydrogen retention observed for DDT68.

3. Hydrogen charging substantially accelerated wear over the investigated load range. The hydrogen-induced wear acceleration factor (I_H/I_0) increased from approximately 2.1 at 200 N to 10.0 at 600 N for DDT68 steel, whereas for P900 the factor varied non-monotonically from approximately 1.4 to 4.2 over the investigated load range. The differences in wear intensity between DDT68 and P900 after hydrogen charging were statistically significant over the investigated load range, demonstrating a substantially stronger hydrogen-induced wear response of DDT68 under the present experimental conditions. This behavior is attributed to the combined effects of hydrogen uptake, microstructural heterogeneity, hydrogen trapping, and localized plastic deformation.
4. Microstructural and fractographic observations revealed damage features associated with localized plastic deformation, subsurface crack nucleation and coalescence, spalling, and delamination wear. Although the SEM observations do not constitute a time-resolved sequence of crack evolution, the observed crack morphology and wear-debris characteristics are consistent with the proposed hydrogen-assisted degradation mechanism and may serve as useful qualitative diagnostic indicators of damage development.
5. Hertzian contact analysis showed that the maximum subsurface shear stresses were located at depths of approximately 0.035–0.062 mm below the contact surface under applied loads of 200–600 N. This depth range defines the critical near-surface region in which hydrogen-assisted crack initiation is expected to develop and provides a rational basis for selecting mechanical surface-strengthening treatments capable of generating compressive residual stresses within the corresponding subsurface layer.
6. A quantitative indicator-based framework was developed for the comparative assessment of hydrogen-assisted degradation by combining the normalized wear response with a hydrogen-related factor. The proposed methodology is intended exclusively for comparative engineering assessment under the investigated experimental conditions and should not be interpreted as a universally calibrated predictive service-life model. The prescribed accumulated mass-loss reference levels provide a consistent numerical basis for comparing the relative degradation behavior of the investigated steels without assigning these levels to specific physical damage events.
7. The proposed indicator-based durability assessment framework provides a physically motivated basis for the comparative evaluation of hydrogen-assisted degradation in rolling–sliding tribological systems and may support the future development of condition-monitoring approaches for hydrogen-related components. The present investigation was conducted under controlled laboratory conditions involving dry rolling–sliding contact, electrochemical hydrogen pre-charging in a 26 wt.% H_2SO_4 solution containing 5 mg/L As_2O_3 , and room-temperature testing. Therefore, the framework should presently be interpreted only within the scope of these experimental conditions. Future research will focus on its validation under high-pressure gaseous hydrogen, elevated operating temperatures, lubricated rolling contact, and service-representative loading conditions. Such investigations will be required before the proposed approach can be extended toward predictive condition monitoring and service-life assessment of hydrogen-related tribological systems.

Author Contributions: Conceptualization, A.I.B., V.O.K., J.C., V.O.B., M.A.K. and L.M.I.; methodology, A.I.B., V.O.K., J.C., V.O.B., M.A.K. and L.M.I.; software, A.I.B., V.O.K., J.C., V.O.B., M.A.K. and L.M.I.; validation, V.O.K., J.C., V.O.B., M.A.K. and L.M.I.; formal analysis, V.O.K., J.C., V.O.B., M.A.K. and L.M.I.; investigation, A.I.B., V.O.K., J.C., V.O.B., M.A.K. and L.M.I.; resources, V.O.K., J.C., V.O.B., M.A.K. and L.M.I.; writing—original draft preparation, A.I.B., V.O.K. and L.M.I.; writing—review and editing, A.I.B., V.O.K. and L.M.I.; visualization, V.O.K. and L.M.I.; project administration, A.I.B.; funding acquisition, A.I.B. All authors have read and agreed to the published version of the manuscript.

Funding: The author (AIB) acknowledges Professor Alicja Krella for inspiration and fruitful discussion during scientific visit at the Institute of Fluid-Flow Machinery Polish Academy of Sciences. The authors acknowledge the Polish Academy of Sciences (PAN, Poland) and the National Academy of Sciences of Ukraine (NANU, Ukraine) for partial support within the framework of the joint Polish–Ukrainian project entitled “Evaluation of operational changes of nanostructure and materials properties in “green” hydrogen systems”, for the period 2025–2027 under the agreement on scientific cooperation between PAN and NANU.

Institutional Review Board Statement: Not applicable.

Informed Consent Statement: Not applicable.

Data Availability Statement: The original contributions presented in this study are included in the article. Further inquiries can be directed to the corresponding author.

Acknowledgments: The authors acknowledge the Polish Academy of Sciences (PAN, Poland) and National Academy of Sciences of Academy of Ukraine (NANU, Ukraine) for partial support in the framework of a joint Polish-Ukrainian projects entitled “Evaluation of operational changes of nanostructure and materials properties in “green” hydrogen systems” for the years 2025–2027 under the agreement on scientific cooperation between the PAN and NANU.

Conflicts of Interest: The authors declare no conflicts of interest.

Abbreviations

The following abbreviations are used in this manuscript:

HNS	High-Nitrogen Steels
HELP	Hydrogen-Enhanced Localized Plasticity
HEDE	Hydrogen-Enhanced Decohesion
RCF	Rolling Contact Fatigue
WEC	White Etching Cracks
WEA	White Etching Areas
NDT	Non-Destructive Testing
XRD	X-ray Diffraction
SMT	Friction Testing Machine (SMT-1)
HRC	Rockwell Hardness (C scale)
Ppm	Parts Per Million
γ -phase	Austenitic phase
γ_N/γ_C	Expanded Austenite (S-phase)
DLC	Diamond-Like Carbon
PVD	Physical Vapor Deposition

References

1. Lynch, S.P. Hydrogen embrittlement phenomena and mechanisms. *Corros. Rev.* **2012**, *30*, 105–123. [[CrossRef](#)]
2. Djukic, M.B.; Sijacki Zeravcic, V.; Bakic, G.M.; Sedmak, A.; Rajicic, B. Hydrogen damage of steels: A case study and hydrogen embrittlement model. *Eng. Fail. Anal.* **2015**, *58*, 485–498. [[CrossRef](#)]
3. Liang, X.Z.; Zhao, G.-H.; Owens, J.; Gong, P.; Rainforth, W.M.; Rivera-Díaz-del-Castillo, P.E.J. Hydrogen-Assisted Microcrack Formation in Bearing Steels under Rolling Contact Fatigue. *Int. J. Fatigue* **2020**, *134*, 105485. [[CrossRef](#)]

4. Ciruna, J.A.; Szieleit, H.J. The Effect of Hydrogen on the Rolling Contact Fatigue Life of AISI 52100 and 440C Steel Balls. *Wear* **1973**, *24*, 107–118. [[CrossRef](#)]
5. Imai, Y.; Endo, T.; Dong, D.; Yamamoto, Y. Study on Rolling Contact Fatigue in Hydrogen Environment at a Contact Pressure below Basic Static Load Capacity. *Tribol. Trans.* **2010**, *53*, 764–770. [[CrossRef](#)]
6. Vegter, R.H.; Slycke, J.T. The Role of Hydrogen on Rolling Contact Fatigue Response of Rolling Element Bearings. *J. ASTM Int.* **2010**, *7*, 1–12. [[CrossRef](#)]
7. Gavriljuk, V.G.; Berns, H. *High Nitrogen Steels: Structure, Properties, Manufacture and Applications*; Springer: Berlin/Heidelberg, Germany, 1999.
8. Chen, Y.-S.; Huang, C.; Liu, P.-Y.; Yen, H.-W.; Niu, R.; Burr, P.; Moore, K.L.; Martínez-Pañeda, E.; Atrons, A.; Cairney, J.M. Hydrogen Trapping and Embrittlement in Metals—A Review. *Int. J. Hydrog. Energy* **2025**, *136*, 789–821. [[CrossRef](#)]
9. Birnbaum, H.K.; Sofronis, P. Hydrogen-Enhanced Localized Plasticity—A Mechanism for Hydrogen-Related Fracture. *Mater. Sci. Eng. A* **1994**, *176*, 191–202. [[CrossRef](#)]
10. Robertson, I.M.; Sofronis, P.; Nagao, A.; Martin, M.L.; Wang, S.; Gross, D.W.; Nygren, K.E. Hydrogen Embrittlement Understood. *Metall. Mater. Trans. A* **2015**, *46*, 2323–2341. [[CrossRef](#)]
11. Krella, A. Hydrogen-Induced Degradation of Metallic Materials—A Short Review. *Materials* **2025**, *18*, 597. [[CrossRef](#)] [[PubMed](#)]
12. Koyama, M.; Akiyama, E.; Tsuzaki, K.; Raabe, D. Hydrogen-Assisted Failure in a Twinning-Induced Plasticity Steel Studied under In Situ Hydrogen Charging by Electron Channeling Contrast Imaging. *Acta Mater.* **2013**, *61*, 4607–4618. [[CrossRef](#)]
13. Balyts'kyi, O.I. Effect of Hydrogen on Structural Strength of High-Nitrogen Chromium–Manganese Steels. *Mater. Sci.* **2000**, *36*, 541–545. [[CrossRef](#)]
14. John, A.; Bąkowski, H.; Śleziak, B. The evaluation of wear processes of the rolling-sliding contact by means of flake wear debris in dry and wet contact. *Multiscale Multidiscip. Model. Exp. Des.* **2024**, *7*, 4965–4977. [[CrossRef](#)]
15. Balyts'kyi, O.I.; Kolesnikov, V.O.; Elias, Y.; Havrylyuk, M.R. Specific Features of the Fracture of Hydrogenated High-Nitrogen Manganese Steels under Conditions of Rolling Friction. *Mater. Sci.* **2015**, *50*, 604–611. [[CrossRef](#)]
16. Balitskii, O.A.; Kolesnikov, V.O.; Balitskii, A.I. Wear Resistance of Hydrogenated High Nitrogen Steel under Dry Friction and Solid-State Lubrication. *Arch. Mater. Sci. Eng.* **2019**, *98*, 57–67. [[CrossRef](#)]
17. Balyts'kyi, O.I.; Kolesnikov, V.O.; Elias, J. Study of the Wear Resistance of High-Nitrogen Steels under Dry Sliding Friction. *Mater. Sci.* **2013**, *48*, 642–646. [[CrossRef](#)]
18. Song, J.; Curtin, W.A. A Nanoscale Mechanism of Hydrogen Embrittlement in Metals. *Acta Mater.* **2011**, *59*, 1557–1569. [[CrossRef](#)]
19. Sofronis, P.; Birnbaum, H.K. Mechanics of the hydrogen–dislocation–impurity interactions—I. Increasing shear modulus. *J. Mech. Phys. Solids* **1995**, *43*, 49–90. [[CrossRef](#)]
20. Michler, T.; Naumann, J. Hydrogen Embrittlement of Cr–Mn–N Austenitic Stainless Steels. *Int. J. Hydrogen Energy* **2010**, *35*, 1485–1492. [[CrossRef](#)]
21. Kim, K.-S.; Kang, J.-H.; Kim, S.-J. Nitrogen Effect on Hydrogen Diffusivity and Hydrogen Embrittlement Behavior in Austenitic Stainless Steels. *Scr. Mater.* **2020**, *184*, 70–73. [[CrossRef](#)]
22. Hojo, T.; Nagasaka, A.; Kobayashi, J.; Shibayama, Y.; Akiyama, E. Effect of Hydrogen on Fatigue Life and Fracture Morphologies of TRIP-Aided Martensitic Steels with Added Nitrogen. *Metals* **2024**, *14*, 346. [[CrossRef](#)]
23. Balitskii, A.I.; Kolesnikov, V.O.; Ivaskevych, L.M.; Balitskii, O.A.; Królikowski, M.A.; Doweiko, J.M. Additively Manufactured Ni–Co Superalloys for Hydrogen Safety Enhancement of Gas-Turbine Energy Systems: Microstructural Degradation and Crack Initiation Mechanisms. *Energies* **2026**, *19*, 3295. [[CrossRef](#)]
24. Fan, Y.H.; Zhang, B.; Wang, J.Q.; Han, E.-H.; Ke, W. Effect of Grain Refinement on the Hydrogen Embrittlement of 304 Austenitic Stainless Steel. *J. Mater. Sci. Technol.* **2019**, *35*, 2213–2219. [[CrossRef](#)]
25. Fan, Y.; Zhang, B.; Wang, J.; Han, E.-H.; Ke, W. Grain Refinement's Effect on Hydrogen Embrittlement of 304 Austenitic Stainless Steel: A Comparative Investigation of Hydrogen In-Situ Charging vs. Pre-Charging. *Int. J. Hydrog. Energy* **2024**, *78*, 22–39. [[CrossRef](#)]
26. Moallemi, M.; Cho, H.-J.; Kim, S.-J. The Adverse Effect of Grain Refinement on Hydrogen Embrittlement in a High Mn Austenitic Steel. *Mater. Sci. Eng. A* **2024**, *916*, 147367. [[CrossRef](#)]
27. Okamoto, R.; Yamamoto, K.; Ono, T.; Macadre, A. Hydrogen Compatibility of SUS304 with Nitrogen Addition and Microstructure Refinement. *J. Jpn. Soc. Heat Treat.* **2024**, *64*, 122–127. [[CrossRef](#)]
28. Luo, Y.; Chen, J. Grain Refinement Alters Deformation Mechanisms to Mitigate Hydrogen Embrittlement in an Interstitial Nitrogen-Alloyed Austenitic Stainless Steel. *Steel Res. Int.* **2026**, *97*, e202500867. [[CrossRef](#)]
29. Zhao, C.; Wu, W.; Deng, J.; Yu, M.; Peng, Y.; Wang, X.; Gong, J. Hydrogen-induced delayed fracture behavior of notched 316L austenitic stainless steel: Role of grain refinement. *Eng. Fail. Anal.* **2024**, *166*, 108880. [[CrossRef](#)]
30. Silverstein, R.; Eliezer, D. Mechanisms of Hydrogen Trapping in Austenitic, Duplex, and Super Martensitic Stainless Steels. *J. Alloys Compd.* **2017**, *720*, 451–459. [[CrossRef](#)]

31. Qin, W.; Thomas, A.; Cheng, Z.Q.; Gu, D.; Li, T.L.; Zhu, W.L.; Szpunar, J.A. Key Factors Affecting Hydrogen Trapping at the Inclusions in Steels: A Combined Study Using Microprint Technique and Theoretical Modeling. *Corros. Sci.* **2022**, *200*, 110239. [CrossRef]
32. López-Uruñuela, F.J.; López-Ortega, A.; Bayón, R.; Pinedo, B.; Aguirrebeitia, J. A Study on the Influence of Electrochemical and Immersion Hydrogen Uptake Methodologies on 100Cr6 Bearing Steel Integrity. *Int. J. Hydrogen Energy* **2025**, *142*, 712–722. [CrossRef]
33. Uyama, H.; Yamada, H.; Hidaka, H.; Mitamura, N. The Effects of Hydrogen on Microstructural Change and Surface Originated Flaking in Rolling Contact Fatigue. *Tribol. Online* **2011**, *6*, 123–132. [CrossRef]
34. Takazaki, D.; Yuya, M.; Neishi, Y.; Kosaka, M.; Sakiyama, Y.; Omura, T.; Kawano, K. Process of White Etching Cracks Formation in Carburized Bearing Steel under Rolling Contact Fatigue. *Tetsu-Hagané* **2024**, *110*, 1150–1164. [CrossRef]
35. Chen, J.-H.; Xie, C.; Li, S.-X.; Lu, S.-Y. White Etching Area Damage Induced by Shear Localization in Rolling Contact Fatigue of Bearing Steel. *Fatigue Fract. Eng. Mater. Struct.* **2024**, *47*, 170–186. [CrossRef]
36. Balitska, V.; Filipecki, J.; Shpotyuk, O.; Swiatek, J.; Vakiv, M. Dynamic Radiation-Induced Effects in Chalcogenide Vitreous Compounds. *Journal of Non-Crystalline Solids* **2021**, *287*, 216–221. [CrossRef]
37. Archard, J.F. Contact and Rubbing of Flat Surfaces. *J. Appl. Phys.* **1953**, *24*, 981–988. [CrossRef]
38. Bhushan, B. *Introduction to Tribology*, 2nd ed.; John Wiley & Sons: New York, NY, USA, 2013; ISBN 978-1-118-40728-3.
39. Sadeghi, F.; Jalahmadi, B.; Slack, T.S.; Raje, N.; Arakere, N.K. A Review of Rolling Contact Fatigue. *J. Tribol.* **2009**, *131*, 041403. [CrossRef]
40. Polonsky, I.A.; Chang, T.P.; Keer, L.M.; Sproul, W.D. An Analysis of the Effect of Hard Coatings on Near-Surface Rolling Contact Fatigue Initiation Induced by Surface Roughness. *Wear* **1997**, *208*, 204–219. [CrossRef]
41. Suh, N.P. The Delamination Theory of Wear. *Wear* **1973**, *25*, 111–124. [CrossRef]
42. Hong, W.; Cai, W.; Wang, S.; Tomovic, M.M. Mechanical Wear Debris Feature, Detection, and Diagnosis: A Review. *Chin. J. Aeronaut.* **2018**, *31*, 867–882. [CrossRef]
43. Kürten, D.; Khader, I.; Raga, R.; Casajús, P.; Winzer, N.; Kailer, A.; Spallek, R.; Scherge, M. Hydrogen Assisted Rolling Contact Fatigue due to Lubricant Degradation and Formation of White Etching Areas. *Eng. Fail. Anal.* **2019**, *99*, 330–342. [CrossRef]
44. Archard, J.F.; Hirst, W. The Wear of Metals under Unlubricated Conditions. *Proc. R. Soc. A* **1956**, *236*, 397–410. [CrossRef]
45. Zaretsky, E.V. *Rolling Bearing Life Prediction, Theory, and Application*; NASA Technical Publication NASA/TP–2013–215305/REV1; Glenn Research Center: Cleveland, OH, USA, 2016; 66p. Available online: <https://ntrs.nasa.gov/api/citations/20160013905/downloads/20160013905.pdf> (accessed on 19 January 2026).
46. Imran, T.; Jacobson, B.; Shariff, A. Quantifying Diffused Hydrogen in AISI 52100 Bearing Steel and in Silver Steel under Tribo-Mechanical Action: Pure Rotating Bending, Sliding–Rotating Bending, Rolling–Rotating Bending and Uni-Axial Tensile Loading. *Wear* **2006**, *261*, 86–95. [CrossRef]
47. Šmelova, V.; Schwedt, A.; Wang, L.; Holweger, W.; Mayer, J. Microstructural Changes in White Etching Cracks (WECs) and Their Relationship with Those in Dark Etching Region (DER) and White Etching Bands (WEBs) due to Rolling Contact Fatigue (RCF). *Int. J. Fatigue* **2017**, *100*, 148–158. [CrossRef]
48. Yin, H.; Wu, Y.; Liu, D.; Zhang, P.; Zhang, G.; Fu, H. Rolling Contact Fatigue-Related Microstructural Alterations in Bearing Steels: A Brief Review. *Metals* **2022**, *12*, 910. [CrossRef]
49. Liang, X.Z.; Rivera-Díaz-del-Castillo, P.E.J. Hydrogen-Accelerated White Etching Area Formation in Bearings under Rolling Contact Fatigue. *Int. J. Fatigue* **2022**, *159*, 106753. [CrossRef]
50. LECO Corporation. *TCH 600-Series (Nitrogen, Oxygen, Hydrogen Determination)–Technical/Application Note*; LECO Corporation: St. Joseph, MI, USA, 2003; 6p. Available online: <https://www.lecomexico.com/inorganico/tch600series209-105.pdf> (accessed on 14 July 2026).
51. Kyriakopoulou, H.P.; Karmiris-Obratański, P.; Tazedakis, A.S.; Daniolos, N.M.; Dourdounis, E.C.; Manolagos, D.E.; Pantelis, D. Investigation of Hydrogen Embrittlement Susceptibility and Fracture Toughness Drop after in situ Hydrogen Cathodic Charging for an X65 Pipeline Steel. *Micromachines* **2020**, *11*, 430. [CrossRef] [PubMed]
52. Michler, T.; San Marchi, C.; Naumann, J.; Weber, S.; Martin, M. Hydrogen Environment Embrittlement of Stable Austenitic Steels. *Int. J. Hydrogen Energy* **2012**, *37*, 16231–16246. [CrossRef]
53. Choi, Y.-H.; Lee, J.-H.; Kim, S.-M.; Lee, D.-H.; Kim, H.-T.; Kim, J.-H.; Kim, M.; Kim, S.-K.; Lee, J.-M. Temperature-Dependent Hydrogen Embrittlement of Austenitic Stainless Steel on Phase Transformation. *Metals* **2023**, *13*, 35. [CrossRef]
54. Wang, Y.; Zhou, Z.; Wu, W.; Gong, J. Warm Pre-Strain: Strengthening the Metastable 304L Austenitic Stainless Steel without Compromising Its Hydrogen Embrittlement Resistance. *Materials* **2017**, *10*, 1331. [CrossRef] [PubMed]
55. Shen, M.; Yan, C.; Bai, Z.; Gong, F.; Li, C.; Li, Q. Comparative Study on Wear and Fatigue Behaviors of Rail Steels with Varying Levels of Martensite and Bainite under Rolling-Sliding Contact Conditions. *Wear* **2025**, *560–561*, 205594. [CrossRef]
56. Singh, P.P.; Ghosh, S.; Mula, S. Characterization of Dry-Sliding Wear Phenomena in a Novel Martensitic Structure Based Low-Carbon Nb and V Alloyed Steel. *Tribol. Int.* **2023**, *189*, 108959. [CrossRef]

57. Li, Q.; Xiao, K.; Lu, Z.; Shi, Y.; Lin, W.; Zhang, S. Mechanism of Laser Carbonitriding Enhancing the Wear Resistance of 45# Steel. *Mater. Today Commun.* **2024**, *38*, 108327. [\[CrossRef\]](#)
58. Balitskii, O.I.; Kolesnikov, V.O.; Ivaskevych, L.M.; Havrylyuk, M.R. The Influence of Specific Features of Load and Hydrogen Charging on Steel. *Tribotech. Prop. Mater. Sci.* **2023**, *58*, 505–512. [\[CrossRef\]](#)
59. Eichinger, M.; Pengg, J.; Zwiitnig, D.; Mori, G.K. Hydrogen Diffusivity, Uptake and Embrittlement of Solution Annealed and Cold Deformed Austenitic Stainless Steel 1.4435 at Hydrogen Pressures up to 1000 Bar and 200 °C. *Corros. Sci.* **2024**, *235*, 112219. [\[CrossRef\]](#)
60. Norouzi, E.; Miresmaeili, R.; Shahverdi, H.R.; Askari-Paykani, M.; Vergani, L.M. Effect of Hydrogen on the Deformation Mechanisms of Metastable 304 Austenitic Stainless Steel. *Mater. Lett.* **2023**, *338*, 134021. [\[CrossRef\]](#)
61. Nie, Y.J.; Yang, F.; Li, X.M.; Li, M.; Wang, Y.Z.; Yin, L.; Shi, Q.X.; Yao, J.; Liang, W.; Zheng, L.W. Hydrogen-Enhanced Densified Deformation Twins in 304L Austenitic Stainless Steel Fabricated by Selective Laser Melting. *Mater. Today Commun.* **2024**, *39*, 109130. [\[CrossRef\]](#)
62. Chmiel, J.; Baranowska, J.; Jędrzejewski, R.; Rzeczycki, A. Stresses in the Surface Layer and Hydrogen Absorption and Diffusion in Cavitation Condition. *Solid State Phenom.* **2015**, *225*, 131–138. [\[CrossRef\]](#)
63. Weihrauch, M.; Patel, M.; Patterson, E.A. Measurements and Predictions of Diffusible Hydrogen Escape and Absorption in Cathodically Charged 316LN Austenitic Stainless Steel. *Sci. Rep.* **2023**, *13*, 10545. [\[CrossRef\]](#) [\[PubMed\]](#)
64. Macadre, A.; Masumura, T.; Manabe, R.; Tsuchiyama, T.; Takaki, S. Effect of Nitrogen-Addition on the Absorption and Diffusivity of Hydrogen in a Stable Austenitic Stainless Steel. *Int. J. Hydrogen Energy* **2019**, *44*, 1263–1271. [\[CrossRef\]](#)
65. Mao, L.Y.; Luo, Z.A.; Huang, C.; Zhou, H.Y.; Zhang, X.M. Exploring the Hydrogen Embrittlement Behavior in Nickel-Economized Austenitic Stainless Steel: Investigating the Role of Manganese in Modifying Hydrogen-Induced Crack Mechanisms. *Corros. Sci.* **2024**, *226*, 111691. [\[CrossRef\]](#)
66. Kolesnikov, V. The Influence of Microstructure Parameters on the Formation of Cutting Products during Machining of Parts from Steels 38KhN3MFA Taking into Account the Influence of Hydrogen. In *Advanced and Novel Technologies—Interdisciplinary Collaboration in Materials Science*; Aikin, M., Smolyakov, O., Girzhon, V., Tkach, D., Belikov, S., Yar-Mukhamedova, G.S., Shalomeev, V., Narivskyi, O., et al., Eds.; ANTICM 2025; Advances in Science, Technology & Innovation; Springer: Cham, Switzerland, 2025; pp. 77–88. [\[CrossRef\]](#)
67. Balitskii, O.A.; Kolesnikov, V.O.; Balitskii, A.I.; Eliaszy, J.J.; Havrylyuk, M.R. Hydrogen effect on the high-nickel surface steel properties during machining and wear with lubricants. *Arch. Mater. Sci. Eng.* **2020**, *104*, 49–57. [\[CrossRef\]](#)
68. Nykyforchyn, H.; Kyryliv, V.; Maksymiv, O. Effect of Nanostructurisation for Structural Steels on Their Wear Hydrogen Embrittlement Resistance. *Solid State Phenom.* **2015**, *225*, 65–70. [\[CrossRef\]](#)
69. Nykyforchyn, H.M.; Lunarska, E.; Kyryliv, V.I.; Maksymiv, O.V. Hydrogen Permeability of the Surface Nanocrystalline Structures of Carbon Steel. *Mater. Sci.* **2015**, *50*, 698–705. [\[CrossRef\]](#)
70. Fernández-Sousa, R.; Betegón, C.; Martínez-Pañeda, E. Analysis of the Influence of Microstructural Traps on Hydrogen-Assisted Fatigue. *Acta Mater.* **2020**, *199*, 253–263. [\[CrossRef\]](#)
71. Koyama, M.; Tasan, C.C.; Nagashima, T.; Akiyama, E.; Raabe, D.; Tsuzaki, K. Hydrogen-Assisted Damage in Austenite/Martensite Dual-Phase Steel. *Philos. Mag. Lett.* **2016**, *96*, 9–18. [\[CrossRef\]](#)
72. Impact of Grinding Media on High-Energy Ball Milling-Driven Amorphization in Multiparticulate As₄S₄/ZnS/Fe₃O₄ Nanocomposites. *Advanced Powder Technology* **2020**, *165*, 722–733. [\[CrossRef\]](#)
73. Chiari, L.; Komatsu, A.; Fujinami, M. Defects Responsible for Hydrogen Embrittlement in Austenitic Stainless Steel 304 by Positron Annihilation Lifetime Spectroscopy. *ISIJ Int.* **2021**, *61*, 1927–1934. [\[CrossRef\]](#)
74. Sugiyama, Y.; Takai, K. Hydrogen Embrittlement Phenomena from Incubation Stage to Fracture Associated with Strain-Induced Vacancy-Hydrogen Complexes in Iron. *ISIJ Int.* **2026**, *66*, 497–507. [\[CrossRef\]](#)
75. Balitskii, A.I.; Syrotyuk, A.M.; Ivaskevich, L.M.; Balitskii, O.A.; Kochmanski, P.; Kolesnikov, V.O. Hydrogen accelerated nanopore nucleation, crack initiation and propagation in the Ni–Co superalloys. *Int. J. Hydrogen Energy* **2024**, *82*, 320–332. [\[CrossRef\]](#)
76. Hembara, O.; Syrotyuk, A.; Chepil, O.; Sapuzhak, Y.; Hembara, N. Evaluation of Increased Local Hydrogen Concentration in the Vicinity of Various Types of Defects in Low-Alloyed Steels. *Procedia Struct. Integr.* **2024**, *59*, 190–197. [\[CrossRef\]](#)
77. Dmytrakh, I.; Syrotyuk, A.; Hembara, O.; Hrynenko, M. Effect of Hydrogen Concentration in Metal on Residual Durability of Defected Pipelines. *Procedia Struct. Integr.* **2024**, *59*, 74–81. [\[CrossRef\]](#)
78. Balitskii, A.A.; Kolesnikov, V.A.; Vus, O.B. Tribotechnical Properties of Nitrogen Manganese Steels under Rolling Friction at Addition of (GaSe)_xIn_{1–x} Powders into Contact Zone. *Metallofiz. Noveishie Tekhnologii* **2010**, *32*, 685–695. Available online: <https://www.scopus.com/pages/publications/77957864676> (accessed on 10 July 2026).
79. Balitskii, O.; Kolesnikov, V. Identification of Wear Products in the Automotive Tribotechnical System Using Computer Vision Methods, Artificial Intelligence and Big Data. In Proceedings of the 2019 XIth International Scientific and Practical Conference on Electronics and Information Technologies (ELIT), Lviv, Ukraine, 16–18 September 2019; pp. 24–27. [\[CrossRef\]](#)

80. Balitskii, A.I.; Kolesnikov, V.O.; Balitska, V.O.; Ivaskevych, L.M.; Dowejko, J.M.; Pilecki, B.J.; Havrilyuk, M.R. Computer-Integrated Surface Image Processing of Hydrogen-Saturated Steel Wear Products. *Appl. Sci.* **2024**, *14*, 11762. [\[CrossRef\]](#)
81. Bucior, M.; Kluz, R.; Trzpieciński, T.; Jurczak, K.; Kubit, A.; Ochał, K. The Effect of Shot Peening on Residual Stress and Surface Roughness of AMS 5504 Stainless Steel Joints Welded Using the TIG Method. *Materials* **2022**, *15*, 8835. [\[CrossRef\]](#) [\[PubMed\]](#)
82. Safyari, M.; Moshtaghi, M. Role of Ultrasonic Shot Peening in Environmental Hydrogen Embrittlement Behavior of 7075-T6 Alloy. *Hydrogen* **2021**, *2*, 377–385. [\[CrossRef\]](#)
83. Pape, F.; Coors, T.; Poll, G. Studies on the Influence of Residual Stresses on the Fatigue Life of Rolling Bearings in Dependence on the Production Processes. *Front. Mech. Eng.* **2020**, *6*, 56. [\[CrossRef\]](#)
84. Saalfeld, S.; Wegener, T.; Scholtes, B.; Niendorf, T. Thermal Stability of Residual Stresses in Differently Deep Rolled Surface Layers of Steel SAE 1045. *J. Mater. Eng. Perform.* **2021**, *30*, 6160–6166. [\[CrossRef\]](#)
85. Nykyforchyn, H.M.; Kyryliv, V.I.; Slobodjan, D.V.; Koval, J.M. Structural steels surface modification by mechanical pulse treatment for corrosion protection and wear resistance. *Surf. Coat. Technol.* **1998**, *100–101*, 125–127. [\[CrossRef\]](#)
86. Celik, O.; Baydogan, M.; Atar, E.; Kayali, E.S.; Cimenoglu, H. Fatigue Performance of Low Temperature Nitrided AISI 321 Grade Austenitic Stainless Steel. *Mater. Sci. Eng. A* **2013**, *565*, 38–43. [\[CrossRef\]](#)
87. Wetegrove, M.; Duarte, M.J.; Taube, K.; Rohloff, M.; Gopalan, H.; Scheu, C.; Dehm, G.; Kruth, A. Preventing Hydrogen Embrittlement: The Role of Barrier Coatings for the Hydrogen Economy. *Hydrogen* **2023**, *4*, 307–322. [\[CrossRef\]](#)
88. Kanjilal, A.; Duarte, M.J.; Scheu, C.; Dehm, G. Hydrogen Permeation Barrier Layers for the Hydrogen Economy. *Annu. Rev. Mater. Res.* **2025**, *55*, 125–149. [\[CrossRef\]](#)
89. Akbari-Kharaji, E.; Shafaie, M.; Sackett, E.; Wood, J.; Djukic, M.B.; Alexander, S. Hydrogen Barrier Coatings: Application and Assessment. *Int. J. Hydrogen Energy* **2025**, *180*, 151666. [\[CrossRef\]](#)
90. De Las Heras, E.; Egidi, D.A.; Corengia, P.; González-Santamaría, D.; García-Luis, A.; Brizuela, M.; López, G.A.; Flores Martinez, M. Duplex Surface Treatment of an AISI 316L Stainless Steel; Microstructure and Tribological Behaviour. *Surf. Coat. Technol.* **2008**, *202*, 2945–2954. [\[CrossRef\]](#)
91. Morita, T.; Andatsu, K.; Hirota, S.; Kumakiri, T.; Ikenaga, M.; Kagaya, C. Effect of Hybrid Surface Treatment Composed of Plasma Nitriding and DLC Coating on Friction Coefficient and Fatigue Strength of Stainless Steel. *Mater. Trans.* **2013**, *54*, 732–737. [\[CrossRef\]](#)
92. Biggio, D.; Elsener, B.; Rossi, A. Ni-P Coatings as Hydrogen Permeation Barriers—A Review. *Coatings* **2025**, *15*, 365. [\[CrossRef\]](#)
93. Wu, L.; Bai, Z.; Hao, Q.; Qin, J. Improving Wear Resistance of DLC-Coated Metal Components During Service: A Review. *Lubricants* **2025**, *13*, 257. [\[CrossRef\]](#)
94. Baur, M.; Khader, I.; Kürten, D.; Schieß, T.; Kailer, A.; Dienwiebel, M. Premature Damage in Bearing Steel in Relation with Residual Stresses and Hydrogen Trapping. *Lubricants* **2024**, *12*, 311. [\[CrossRef\]](#)
95. Holweger, W.M.; Gegner, J. Rolling Contact Fatigue and White Etching Cracks of Bearings. *Lubricants* **2023**, *11*, 247. [\[CrossRef\]](#)
96. Holubets, V.M.; Dovhunyk, V.M.; Pashechko, M.I.; Kornii, S.A.; Shpulyar, Y.S. Friction Behavior of Electric-Spark Coatings under the Conditions of Boundary Lubrication. *Mater. Sci.* **2020**, *56*, 43–49. [\[CrossRef\]](#)
97. Halaichak, S.A.; Korniy, S.A.; Zakiev, V.I.; Vynar, V.A.; Danylchuk, M.V.; Chuchman, M.R.; Mardarevych, R.S. Corrosion Resistance, Wear Resistance, and Tribocorrosion of Ni–Mo–B Electrodeposited Coatings. *Mater. Sci.* **2025**, *61*, 580–589. [\[CrossRef\]](#)
98. Balyts'kyi, O.I.; Kolesnikov, V.O.; Kawiak, P. Triboengineering Properties of Austenitic Manganese Steels and Cast Irons under Sliding Friction Conditions. *Mater. Sci.* **2005**, *41*, 624–630. [\[CrossRef\]](#)
99. Kindrachuk, M.; Volchenko, D.; Balitskii, A.; Abramek, K.F.; Volchenko, M.; Balitskii, O.; Skrypnyk, V.; Zhuravlev, D.; Yurchuk, A.; Kolesnikov, V. Wear Resistance of Spark Ignition Engine Piston Rings in Hydrogen-Containing Environments. *Energies* **2021**, *14*, 4801. [\[CrossRef\]](#)
100. Balitskii, A.; Kolesnikov, V.; Abramek, K.F.; Balitskii, O.; Elias, J.; Marya, H.; Ivaskevych, L.; Kolesnikova, I. Influence of Hydrogen-Containing Fuels and Environmentally Friendly Lubricating Coolant on Nitrogen Steels' Wear Resistance for Spark Ignition Engine Pistons and Rings Kit Gasket Set. *Energies* **2021**, *14*, 7583. [\[CrossRef\]](#)
101. Enami, K.; Yamada, H.; Komata, H. Hydrogen Generation from Lubricant under Rolling-Sliding Contact. *Tribol. Online* **2024**, *19*, 298–307. [\[CrossRef\]](#)
102. Ruellan, A.; Stadler, K.; Jelita Rydel, J.; Ryan, H. The Influence of Lubricant Formulation on Early Thrust and Radial Bearing Damage Associated with White Etching Cracks. *Proc. Inst. Mech. Eng. Part J. J. Eng. Tribol.* **2021**, *235*, 1047–1059. [\[CrossRef\]](#)
103. Linto, D.; Ramkumar, P. Lubricant Performance against White Etching Areas (WEAs) Formation in AISI 52100 Bearing Steel under Cyclic Compressive Loading. *Proc. Inst. Mech. Eng. Part J. J. Eng. Tribol.* **2023**, *237*, 1012–1024. [\[CrossRef\]](#)
104. Tanaka, H.; Ratoi, M.; Sugimura, J. The Role of Synthetic Oils in Controlling Hydrogen Permeation of Rolling/Sliding Contacts. *RSC Adv.* **2021**, *11*, 726–738. [\[CrossRef\]](#) [\[PubMed\]](#)
105. Han, W.; Mu, X.; Liu, Y.; Wang, X.; Li, W.; Bai, C.; Zhang, H. A Critical Review of On-Line Oil Wear Debris Particle Detection Sensors. *J. Mar. Sci. Eng.* **2023**, *11*, 2363. [\[CrossRef\]](#)

106. Zhang, H.; Shi, H.; Li, W.; Ma, L.; Zhao, X.; Xu, Z.; Wang, C.; Xie, Y.; Zhang, Y. A Novel Impedance Micro-Sensor for Metal Debris Monitoring of Hydraulic Oil. *Micromachines* **2021**, *12*, 150. [[CrossRef](#)] [[PubMed](#)]
107. Balyts'kyi, O.I.; Kolesnikov, V.O. Investigation of Wear Products of High-Nitrogen Manganese Steels. *Mater. Sci.* **2009**, *45*, 576–581. [[CrossRef](#)]
108. Liu, X.; Wang, J.; Sun, K.; Cheng, L.; Wu, M.; Wang, X. Semantic Segmentation of Ferrography Images for Automatic Wear Particle Analysis. *Eng. Fail. Anal.* **2021**, *122*, 105268. [[CrossRef](#)]
109. Wang, S.; Shao, T.; Wu, T.; Sarkodie-Gyan, T.; Lei, Y. Knowledge-Guided Convolutional Neural Network Model for Similar Three-Dimensional Wear Debris Identification with Small Number of Samples. *J. Tribol.* **2023**, *145*, 091105. [[CrossRef](#)]
110. Kosarevych, R.Y.; Student, O.Z.; Svirs'ka, L.M.; Rusyn, B.P.; Nykyforchyn, H.M. Computer Analysis of Characteristic Elements of Fractographic Images. *Mater. Sci.* **2013**, *48*, 474–481. [[CrossRef](#)]
111. Rusyn, B.P.; Ivanyuk, V.H.; Kapshii, O.V.; Kornii, V.V. Evaluation of the Characteristics of Cracks According to the Images of the Surfaces of Materials. *Mater. Sci.* **2007**, *43*, 574–579. [[CrossRef](#)]
112. Glotka, A.A.; Moroz, A.N. Comparison of the Effects of Carbides and Nonmetallic Inclusions on Formation of Fatigue Microcracks in Steels. *Met. Sci. Heat Treat.* **2019**, *61*, 521–524. [[CrossRef](#)]
113. Moroz, A.N.; Glotka, A.A. Relation Between the Structural Components and Nonmetallic Inclusions and the Operating Properties of Bearing Steel. *Met. Sci. Heat Treat.* **2018**, *59*, 575–578. [[CrossRef](#)]
114. Tkachov, V.I.; Ivas'kevych, L.M.; Vytvyts'kyi, V.I. Methodological aspects of determination of hydrogen resistance of steels. Methodological aspects of determination of hydrogen resistance of steels. *Mater. Sci.* **2002**, *38*, 484–493. [[CrossRef](#)]
115. Tkachov, V.I.; Ivaskevych, L.M.; Mochulskyi, V.M. Temperature dependences of the mechanical properties of austenitic and martensitic steels in hydrogen. *Mater. Sci.* **2007**, *43*, 654–666. [[CrossRef](#)]
116. Pohrelyuk, I.M.; Fedirko, V.M.; Lavrys, S.M. Effect of Preliminary Ball Burnishing on Wear Resistance of the Nitrided VT22 Alloy. *J. Frict. Wear* **2017**, *38*, 221–224. [[CrossRef](#)]
117. Anoop, C.R.; Singh, R.K.; Kumar, R.R.; Miyala, J.; Murty, S.V.S.N.; Tharian, K.T. Development and Validation of Processing Maps for Hot Deformation of Modified AISI 321 Austenitic Stainless Steel. *Mater. Perform. Charact.* **2020**, *9*, 150–169. [[CrossRef](#)]
118. Balitskii, A.; Krohmalny, O.; Ripey, I. Hydrogen cooling of turbogenerators and the problem of rotor retaining ring materials degradation. *Int. J. Hydrogen Energy* **2000**, *25*, 167–171. [[CrossRef](#)]
119. Balitskii, A.I.; Ivaskevich, L.M.; Balitskii, O.A. Rotor Steels Crack Resistance and Fracture Behavior for Hydrogen Targeted Materials Ever-Widening Database. *Eng. Fract. Mech.* **2022**, *260*, 108168. [[CrossRef](#)]
120. Tretiak, O.; Kravchenko, S.; Shestak, B.; Shpitalnyi, D.; Arefieva, M.; Tretiak, I.; Serhiienko, S.; Kovryga, A. Devising a Method for Designing Multicomponent Diffusers of Compressors in Turbogenerators with Hydrogen Cooling. *East.-Eur. J. Enterp. Technol.* **2025**, *4*, 26–38. [[CrossRef](#)]
121. Smetankina, N.; Morhun, S. Determination of the Modern Marine Single Shaft Gas Turbine Rotor Blades Fatigue Strength Parameters. *Int. J. Turbo Jet.-Engines* **2024**, *42*, 271–276. [[CrossRef](#)]
122. Trzepieciński, T. Modeling the Friction Behavior of Low-Carbon Steel Sheets Using Various Machine Learning Algorithms Based on Strip Drawing Test Data. *Materials* **2026**, *19*, 1109. [[CrossRef](#)] [[PubMed](#)]
123. Tolmachov, V.; Riabko, A.; Hrudynin, B.; Marynchenko, Y.; Rozhkova, A.; Ihnatieva, V. Determination of Flax Fiber Quality Indicators Taking into Account Sound-Absorbing Properties for Robotic Landscaping Systems. *Fibres Text.* **2026**, *33*, 27–40. [[CrossRef](#)]
124. Prokhorova, V.V.; Yemelyanov, O.Y.; Koleschchuk, O.Y.; Antonenko, N.S.; Zaitseva, A.S. Information Support for Management of Energy-Saving Economic Development of EnterprisesNauk. *Visnyk Natsionalnoho Hirnychoho Universytetu* **2023**, *6*, 175–176. [[CrossRef](#)]

Disclaimer/Publisher's Note: The statements, opinions and data contained in all publications are solely those of the individual author(s) and contributor(s) and not of MDPI and/or the editor(s). MDPI and/or the editor(s) disclaim responsibility for any injury to people or property resulting from any ideas, methods, instructions or products referred to in the content.