

Article

Evolutionary Relationship Between Microstructure and Hydrogen Distribution During the Tensile of Pipeline Girth Welds

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Abstract

Hydrogen trapping is a key mechanism to mitigate hydrogen embrittlement (HE) in pipeline steels, yet how trapping behavior evolves during plastic deformation and which microstructural features are responsible remain poorly understood. In this work, interrupted slow strain rate tensile tests were performed on pipeline girth weld metal (WM) under 6.3 MPa H₂ combined with thermal desorption spectroscopy (TDS), hydrogen microprint technique (HMT), and electron backscatter diffraction (EBSD). The results reveal that hydrogen trapping is highly strain-dependent. In the undeformed stage, only weak traps (Peak A at ~250 °C) and moderate traps (Peak B at ~500 °C) are present, as identified by TDS. Upon tensile straining to 15%, a new strong trap (Peak C at ~700 °C) emerges and intensifies, which is attributed to deformation-induced defects within acicular ferrite (AF). HMT shows that hydrogen distribution becomes homogeneous inside AF grains with increasing strain, while remaining localized at grain boundaries in proeutectoid ferrite (PF). EBSD indicates that AF undergoes more uniform plastic deformation and dislocation accumulation than PF. Most importantly, the WM with a higher AF content (74.5%) exhibits a significantly lower HE index (13.4%) than the WM with a lower AF content (53.9%, HE index 23.5%), confirming the beneficial role of AF. These findings demonstrate that AF is not merely a static hydrogen trap but a dynamic source of strain-induced strong traps, which effectively immobilize hydrogen and reduce embrittlement susceptibility under service-relevant deformation.

Keywords: pipeline girth weld; hydrogen embrittlement; acicular ferrite; strain-dependent hydrogen trapping; thermal desorption spectroscopy



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1. Introduction

Hydrogen embrittlement (HE) poses a major challenge to the safe transportation of hydrogen through steel pipelines [1–3]. Girth welds are particularly vulnerable due to their heterogeneous microstructure and residual stresses [4,5]. One promising mitigation strategy is to design microstructures that contain effective hydrogen traps, thereby reducing the concentration of diffusible hydrogen that can reach crack tips [6,7].

Acicular ferrite (AF)—a fine-grained, interlocking intragranular ferrite—has long been considered an ideal microstructure for HE resistance in pipeline steels [8–10]. Studies have

shown that AF exhibits high hydrogen trapping efficiency and high resistance to hydrogen-induced cracking [11,12]. The conventional explanation focuses on its high density of high-angle grain boundaries (HAGBs) and fine grain size, which act as static traps and impede crack propagation [13–18]. However, these studies mainly focused on the effect of AF on hydrogen diffusion and hydrogen distribution under the undeformed stage and ignored its critical role during the deformation process.

However, in service, pipelines undergo local plastic deformation due to internal pressure fluctuations, thermal gradients, and ground settlement [19–25]. Under such dynamic conditions, the microstructure evolves. Dislocations multiply, grain boundaries rotate, and new defects form [26–30]. It remains unclear how hydrogen trapping behavior changes with increasing strain, whether the traps that exist in undeformed AF remain dominant after plastic deformation, and whether AF is still beneficial under these dynamic conditions [31–35].

Current understanding is largely based on static or pre-deformed specimens, which do not capture the co-evolution of hydrogen distribution and microstructure during active straining [36–41]. To address this gap, interrupted slow strain rate tensile (SSRT) tests were performed on pipeline girth weld metal in high-pressure hydrogen, and hydrogen desorption, hydrogen distribution, and microstructure were characterized at multiple strain levels. The aims are: (i) to reveal the strain-dependent evolution of hydrogen traps, (ii) to identify which microstructural constituents contribute to the deformation-induced traps, and (iii) to re-evaluate the role of AF content in HE susceptibility under dynamic loading.

2. Materials and Methods

In this study, an L245 pipe with an outer diameter of 610 mm and a wall thickness of 15.9 mm was selected as the welding base metal. A U-shaped groove was machined at the pipe end. Gas metal arc welding (GMAW) was employed using a shielding gas composed of 80% Ar and 20% CO₂. This type of shielding gas is beneficial for improving arc stability, reducing welding spatter, ensuring weld formation and cutting down the cost. An ER70S-G solid wire conforming to the AWS A5.18 standard [42] was used as the welding consumable, with its chemical composition presented in Table 1. The welding process consisted of 7 layers and 8 passes. The AF content in the weld metal (WM) was varied by adjusting the welding process parameters. Meanwhile, the welding process parameters can guarantee adaptability to factors such as on-site assembly quality, groove machining quality and defect control. Detailed welding process parameters are listed in Table 2.

Table 1. Chemical composition of welding wire.

C	Si	Mn	P	S	Cr	Mo	Ni	Cu	Al	Ti	Fe
0.07	0.75	1.54	0.012	0.008	0.035	0.008	0.03	0.13	0.003	0.06	Bal.

Table 2. Welding process parameters.

No.	Welding Current (A)	Welding Voltage (V)	Welding Speed (mm/min)	Heat Input (kJ/mm)
1	133–147	22.6–23.4	220–270	0.74–0.89
2	155–156	21.9–22.7	300–310	0.68–0.71

Slow strain rate tensile (SSRT) tests were conducted to evaluate the HE susceptibility of two WMs with different AF contents. Tensile specimens were machined from pipeline girth

welds, with their longitudinal direction aligned with the circumferential direction of the weld. The specimens were sampled near the 3 o'clock position, featuring a gauge length of 30 mm and a diameter of 6 mm, as shown in Figure 1. After machining, the specimens were meticulously ground along their axial direction using 2000# fine-grit sandpaper (SHARP-NESS; HuBei Yuli Abrasive Belts Group Co., Ltd., Tongcheng, China). Subsequently, the specimens were cleaned in a solution mixed with acetone (analytical grades, Sinopharm Chemical Reagent Co., Ltd., Shanghai, China) and alcohol (analytical grades, Sinopharm Chemical Reagent Co., Ltd., Shanghai, China) to completely remove grease or contaminants. A slow strain rate of 1×10^{-5} /s was adopted. Tensile tests were conducted under both 6.3 MPa N₂ and 6.3 MPa H₂ atmospheres, with a gas purity of 99.999% for both environments. The SSRT testing procedure was carried out in accordance with the ASTM G129 standard [43] practice for slow strain rate testing to evaluate the susceptibility of metallic materials to environmentally assisted cracking. For each combination of WM and testing atmosphere, three repeated SSRT tests were performed.

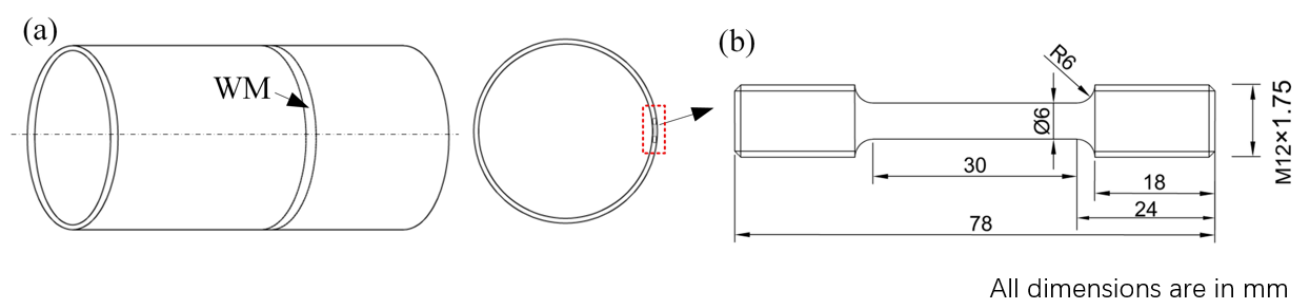


Figure 1. SSRT specimens' position in the girth weld (a) and dimension (b).

Interrupted deformation experiments were performed on girth weld specimens obtained using welding procedure parameter No. 2, as listed in Table 2. Three strain levels were selected: 0% strain (undeformed stage, prior to loading), 10% strain (pre-necking, within the uniform deformation stage), and 15% strain (post-necking, within the non-uniform deformation stage). In addition, the specimen that fractured completely during tensile testing was also characterized. Specifically, WM specimens were pre-charged with hydrogen under a 6.3 MPa hydrogen atmosphere for at least 48 h, then loaded to the target strain levels, and subsequently unloaded. Afterwards, characterization tests including electron backscatter diffraction (EBSD), hydrogen microprint technique (HMT), and thermal desorption spectroscopy (TDS) were conducted on the specimens at different strain levels.

For the HMT test, a self-prepared AgBr photosensitive emulsion was used. A uniform monolayer of the AgBr photosensitive film was applied onto the surface of the hydrogen-charged specimen using the wire-winding method. After the specimen was allowed to stand in a darkroom to enable the reaction between hydrogen and the film, the unreacted AgBr was removed with a sodium thiosulfate solution. The distribution of Ag indirectly reflects the distribution of H. Finally, the distribution of silver in the metal microstructure was observed using a JEOL JSM-7200F scanning electron microscope (SEM) (JEOL Ltd., Tokyo, Japan) equipped with an Oxford X-MAX50 energy-dispersive X-ray spectrometer (EDS) (JEOL Ltd., Tokyo, Japan).

Prior to EBSD, the specimens were electropolished in a perchloric acid alcohol solution (8 vol%) at an electrolysis voltage of 30 V for 24 s. EBSD characterization was performed using a JSM-7200F SEM equipped with an EBSD detector (Oxford Instruments, Oxford, UK), operating at 20 kV, 70° tilt, and a step size of 0.2 μ m. Data analysis was carried out using Channel 5 software (Version number 5.0.9.0) to obtain band contrast (BC) maps, inverse pole figure (IPF) maps, grain boundary maps, Kernel average misorientation (KAM)

maps, and geometrically necessary dislocation (GND) density calculations. A maximum misorientation angle of 5° was used for KAM analysis.

For TDS, the specimens were heated at a rate of $200\text{ }^\circ\text{C/h}$ over a temperature range from $30\text{ }^\circ\text{C}$ to $900\text{ }^\circ\text{C}$ and held at the maximum temperature for 10 min. In addition to the interrupted deformation specimens, TDS measurements were also performed on the fractured tensile specimens of both WMs after testing to fracture at 6.3 MPa H_2 under the same SSRT conditions. These specimens were used to compare the hydrogen trapping characteristics between the AF-rich and AF-poor WMs at the final stage of deformation.

3. Results

3.1. Macroscopic HE Susceptibility of WMs with Different AF Contents

By adjusting the welding process parameters, two types of weld microstructures were obtained. The representative microstructures obtained with process parameters 1 and 2 (Table 2) are shown in Figure 2a,c, respectively, both of which consist of intragranular AF (dark gray) and intergranular proeutectoid ferrite (PF, light gray). The PF regions in the images were stained yellow using the image processing software ImageJ (Version number 1.46), and their area fraction in the image was statistically calculated. The staining results are presented in Figure 2b,d. Statistical analysis of the area fraction of the stained regions was performed, which reflects the PF content in the WM, with the remaining balance considered AF. It should be noted here that a certain amount of Widmanstätten ferrite (WF) is also present in the WM. In the statistics, WF is classified into the PF range. At least 30 microstructure maps were statistically recorded for each WM. The statistical results show that the AF contents of the two different WMs are 74.50% and 53.90%, respectively.

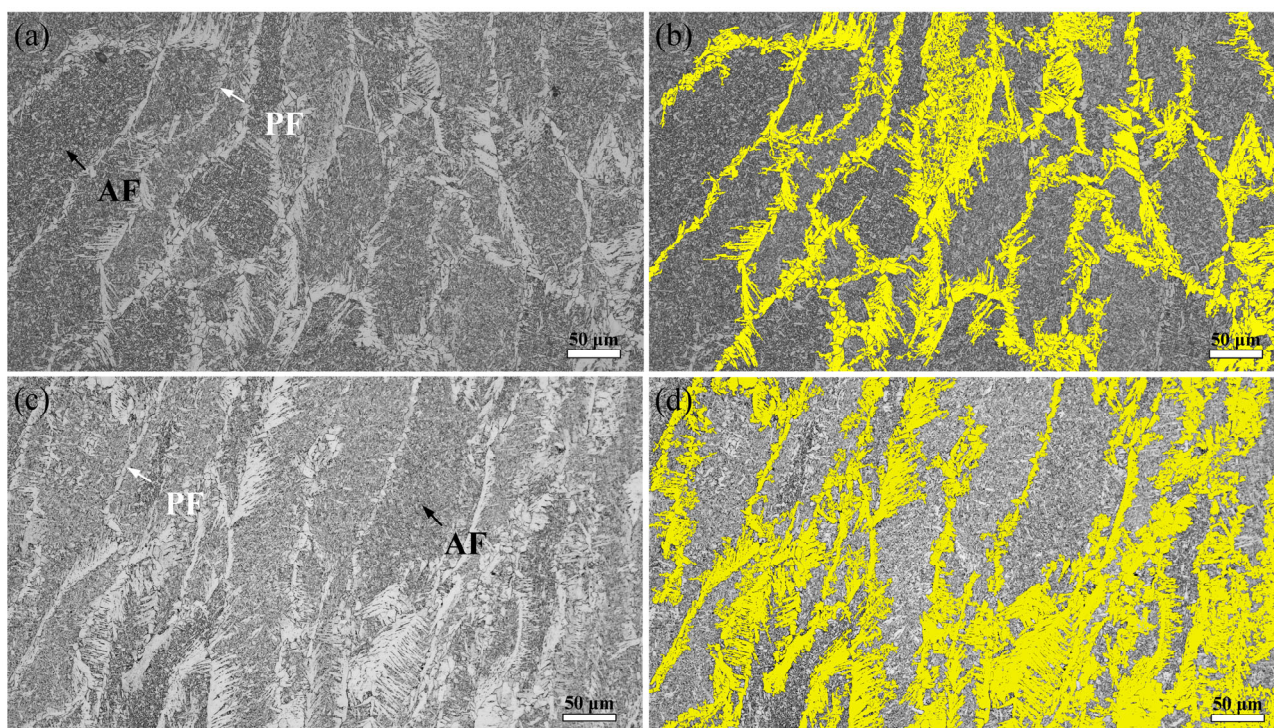


Figure 2. Microstructures before and after PF staining at different AF contents. (a,b) ~74.50% AF; (c,d) ~53.90% AF; (a,c) original OM (un-stained); (b,d) PF stained in yellow.

The representative stress–strain curves for different WMs in N_2 and H_2 atmospheres are shown in Figure 3a. For both types of WM, the yield strength and tensile strength are essentially the same in both N_2 and H_2 gas environments. However, hydrogen significantly

reduces the total strain of the WM (reflecting the overall deformation capability), and the extent of this reduction depends on the WM type: the hydrogen-induced degradation of strain capability is markedly smaller for the WM with an AF content of ~74.50% than for that with ~53.90% AF.

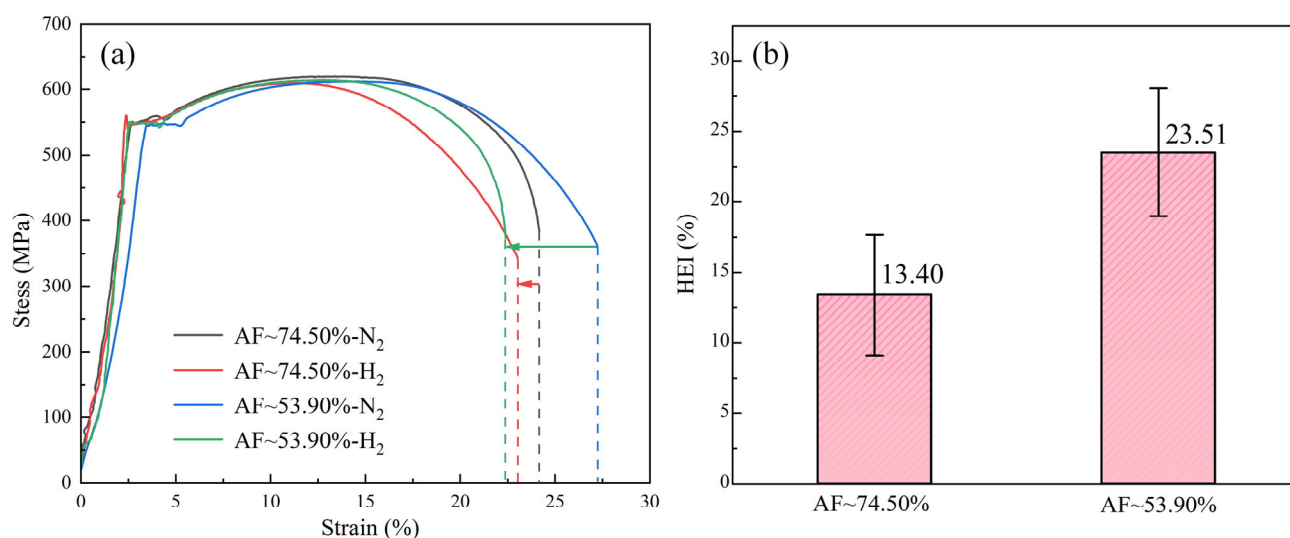


Figure 3. Representative stress–strain curves in N₂ and H₂ (a) and HEI (b) for WMs with varying AF contents. In (a), dashed lines mark the total strain at fracture, and leftward arrows show the H₂-induced decrease in total strain.

The reduction in area reflects the plastic deformation capability of a material; a higher reduction in area (Ψ) indicates a stronger plastic deformation ability. The loss in Ψ caused by hydrogen is commonly used to quantitatively evaluate the HE susceptibility of a material:

$$\text{HEI} = \frac{\Psi_{N_2} - \Psi_{H_2}}{\Psi_{N_2}} \times 100\% \quad (1)$$

where HEI is the HE index, and Ψ_{N_2} and Ψ_{H_2} are the reductions in areas of the same WM tested in N₂ and H₂, respectively. The calculated HE index for the two WMs with different AF contents tested in 6.3 MPa H₂ (relative to the N₂ environment) is presented in Figure 3b. Clearly, the HEI at an AF content of approximately 74.5% (13.40%) is lower than that at an AF content of approximately 53.9% (23.51%). These results indicate that a higher AF content is beneficial for reducing the HE susceptibility of the WM.

3.2. Strain-Dependent Microstructure, Hydrogen Distribution, and Trapping Behavior in the WM

3.2.1. Strain-Dependent Microstructure Evolution

Figure 4a–c show the BC maps with different deformation magnitudes. It can be observed that the grain size of PF remains relatively unchanged with increasing strain. However, the grain size of AF gradually decreased from 4.8 μm in the undeformed stage to 3.2 μm under uniform plastic deformation, and further to 2.1 μm under non-uniform plastic deformation. This indicates that increased plastic deformation led to the fragmentation of AF grains.

Figure 4d–i show the IPF maps with grain boundary angles and the statistical frequency of local misorientation at grain boundaries (GBs), respectively. The grain boundary angles are classified into low-angle grain boundaries (LAGBs, $3^\circ \leq \text{LAGBs} < 15^\circ$, blue lines), medium-angle grain boundaries (MAGBs, $15^\circ \leq \text{MAGBs} < 45^\circ$, green lines), and high-angle grain boundaries (HAGBs, $45^\circ \leq \text{HAGBs}$, red lines). It can be observed that the GBs of PF are mainly low-angle grain boundaries, whereas those of AF predominantly

consist of MAGBs and HAGBs. This indicates that AF contributes the majority of HAGBs. As the strain magnitude increases, newly formed and discontinuous LAGBs appear in local regions within both PF and AF grains, indicating significant plastic deformation in these areas. Furthermore, with increasing strain magnitude, the proportion of sub-angle grain boundaries (SAGBs < 3°) gradually rises from 45.45% to 58.61% and then to 68.74%. In contrast, the proportion of HAGBs gradually decreases from 32.14% to 23.66% and then to 15.58%. It should be noted that this does not imply a reduction in the total density of HAGBs; rather, it reflects a substantial increase in local dislocation density, leading to the formation of more SAGBs.

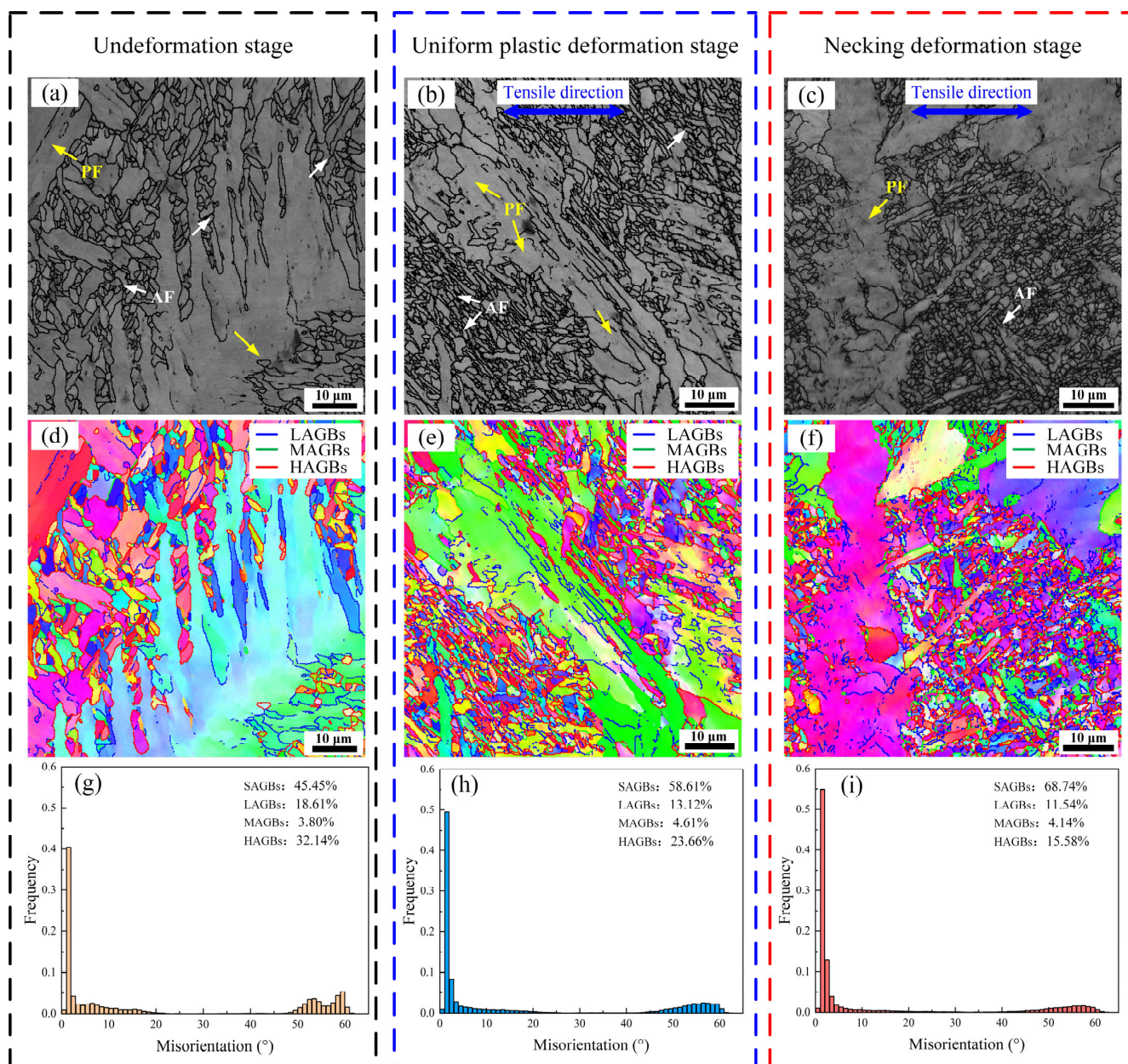


Figure 4. Microstructure characterization with different deformation magnitudes. (a–c) BC maps; (d–f) IPF; (g–i) statistical frequency of local misorientation.

Figure 5 shows the distributions of KAM and GNDs at different strain magnitudes. In the undeformed stage, higher KAM values are primarily concentrated at low-angle grain boundaries (LAGBs) of PF, as indicated by white arrows. As the strain magnitude increases, KAM rises inside both PF grains (red arrows) and AF grains (yellow arrows).

Moreover, the KAM distribution inside AF grains is more uniform, whereas localized KAM concentration occurs inside PF grains. In other words, strain is evenly distributed within AF grains, while local strain concentration occurs in PF grains. This could be primarily attributed to the fine and densely distributed AF grains, which preferentially facilitate coordinated plastic deformation. In contrast, coarse PF grains surrounded by AF deform with difficulty in a coordinated manner. Figure 5d–f reveal that as the deformation level increases, dislocation density gradually rises from 5.96 to 10.43 and then to 12.65 (unit: $1.0 \times 10^{14} \text{ m}^{-3}$). Meanwhile, dislocations are predominantly distributed in the AF region. In the undeformed stage, a certain degree of dislocation concentration exists at local grain boundaries, as shown by magenta arrows in Figure 5g. As the deformation extent increases, the dislocation density at the grain boundaries of PF increases slightly, as indicated by the white arrows in Figure 5h. It should be noted that the dislocation density at the boundaries and inside AF grains (including fragmented grains) increases significantly.

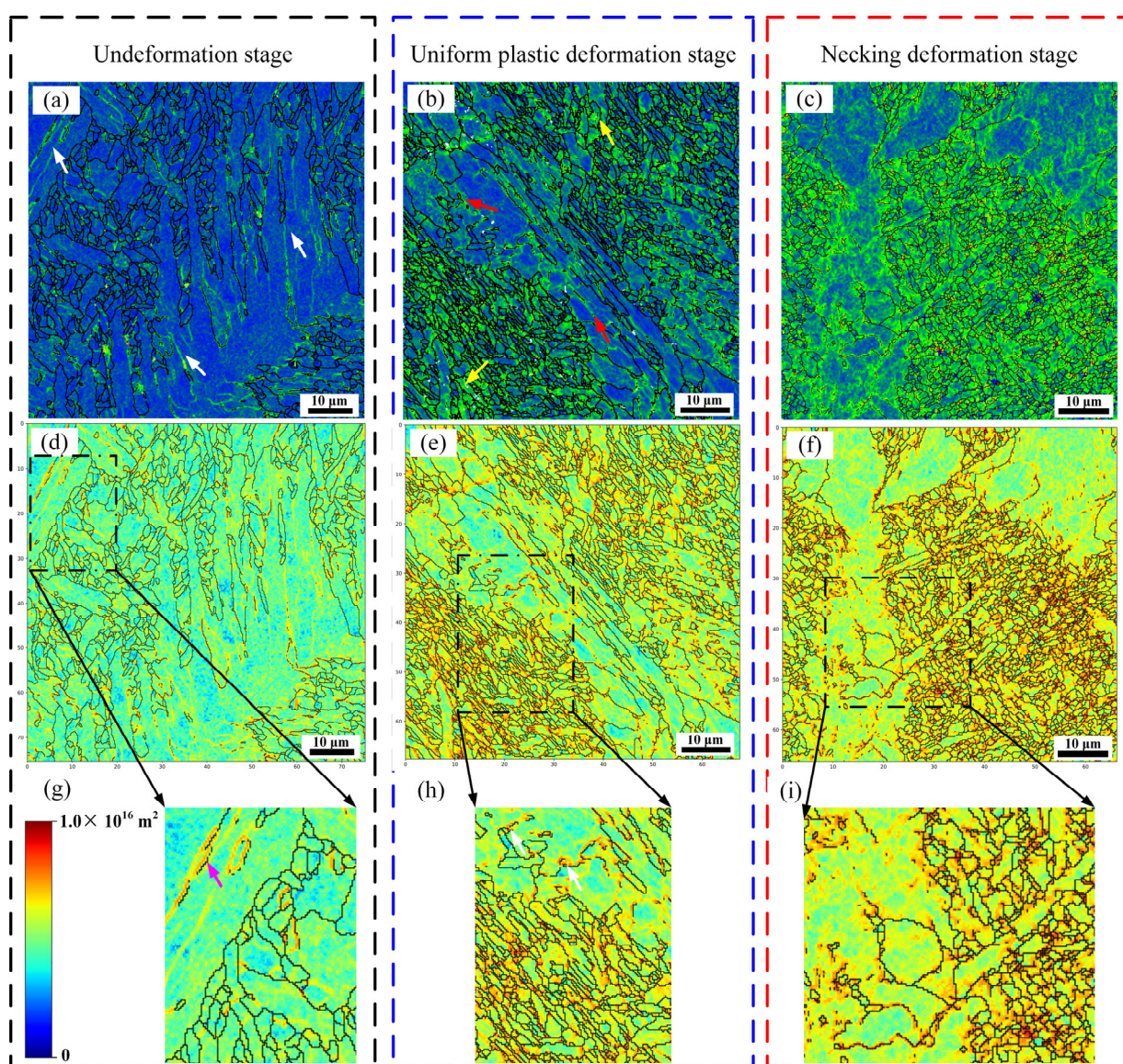


Figure 5. Distributions of both KAM and GNDs with different strain magnitudes. (a–c) KAM maps; (d–i) GND maps.

3.2.2. Strain-Dependent Hydrogen Distribution (HMT)

Figure 6 shows the distribution of both the microstructure and Ag element (as a tracer for hydrogen) with different strain magnitudes. During the undeformed stage, H atoms are primarily distributed at the grain boundaries of both PF and AF, as indicated by the red and white arrows in Figure 6d, respectively. It is noteworthy that this situation is basically consistent with the strain distribution of PF and distribution of AF in HAGBs. This indicates that strain concentration locations and grain boundaries serve as the initial hydrogen trapping sites. As the deformation magnitude increases to the uniform plastic deformation stage, besides the H atom aggregation at grain boundaries, the remaining H atoms are distributed relatively evenly inside the grains of both PF and AF. This could be primarily attributed to the plastic deformation occurring inside the grains of these two microstructures, leading to an increase in lattice defects such as dislocations. In turn, more H atoms can be trapped. As the deformation magnitude increases to the non-uniform plastic deformation stage, the H atoms distribution becomes more uniform in the AF region, as shown in the white box in Figure 6f. This is mainly because AF is more prone to coordinated plastic deformation. As a result, the AF grains underwent fragmentation, leading to the formation of new and finer grains, as shown in Figures 4f and 5f. In addition, for PF, the H atoms remain concentrated at the grain boundaries of PF, as indicated by the red arrows in Figure 5f, which can be ascribed to the fact that the strain of PF remains primarily distributed at grain boundaries, as indicated by the white arrows in Figure 5h,i.

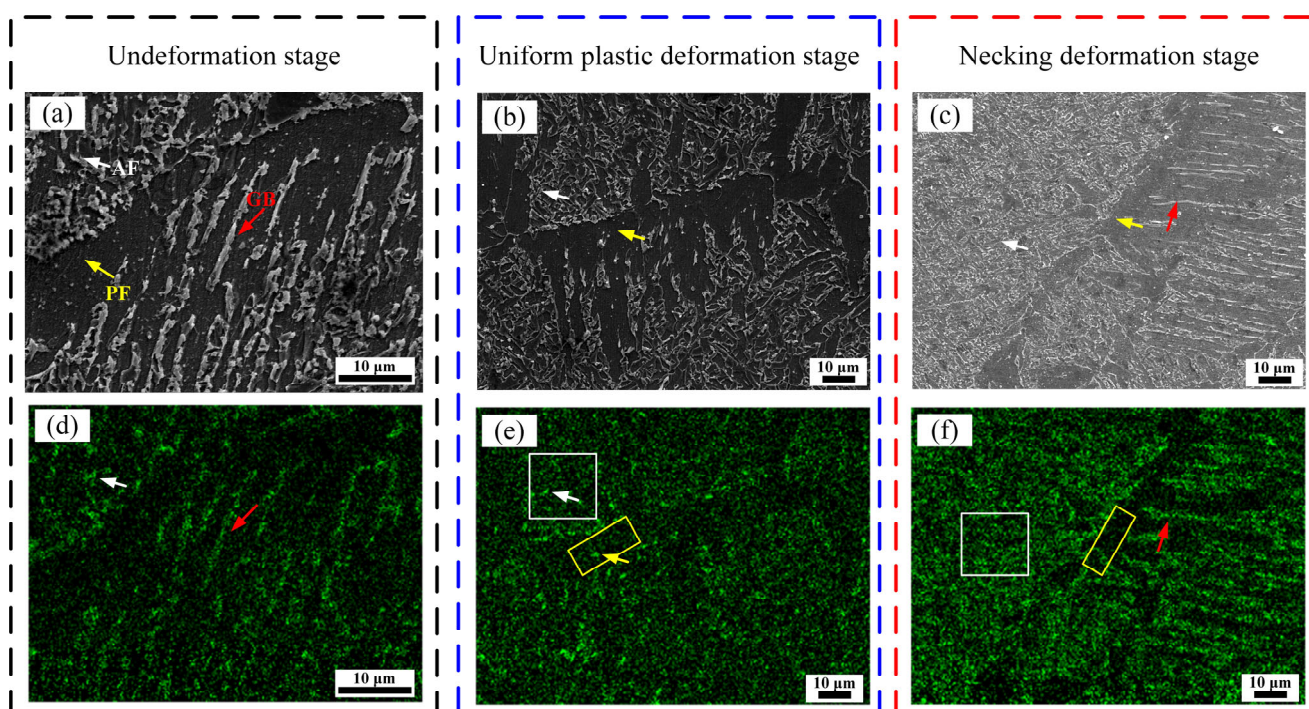


Figure 6. Distributions of both microstructure and Ag element with different strain magnitudes. (a–c) Microstructure under SEM; (d–f) Ag element distribution. In (e,f), white boxes enclose acicular ferrite (AF), and yellow boxes mark proeutectoid ferrite.

3.2.3. Strain-Dependent Hydrogen Traps (TDS)

Figure 7 shows the TDS results at the different deformation stages. Figure 7a shows hydrogen desorption at the undeformed stage. It can be observed that a hydrogen desorption peak (Peak A1) appears at approximately 250 °C, and another hydrogen desorption peak (Peak B1) emerges at around 500 °C. Peak A1 corresponds to some weak hydrogen traps, while Peak B1 is a mixture of other weak hydrogen traps (such as HAGBs) and strong

hydrogen traps (such as precipitates). The HAGBs mainly originate from the AF, as shown in Figure 4d–i. And the precipitates mainly come from the heterogeneous nucleation of the AF, such as Ti(C,N) [44–46]. As a result, H atoms are distributed at the grain boundaries of AF and PF under the undeformed stage, as indicated by the white and red arrows in Figure 4a.

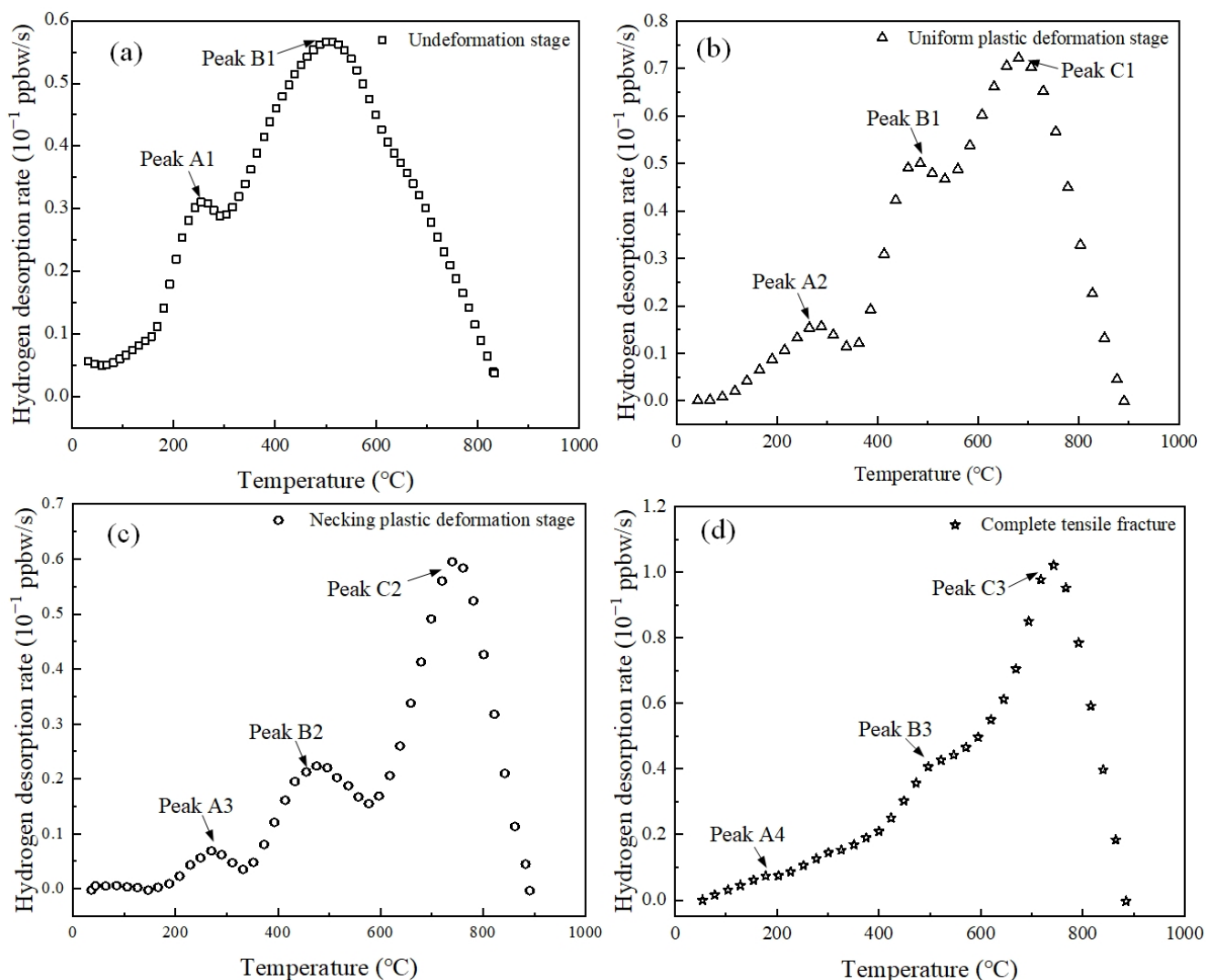


Figure 7. TDS results for different strain magnitudes: (a) undeformation stage; (b) uniform plastic deformation stage; (c) non-uniform deformation stage; (d) complete tensile fracture.

Figure 7b shows the hydrogen desorption behavior at the stage of uniform plastic deformation. It can be observed that there are three hydrogen desorption peaks, namely, one near 250 °C (Peak A2), one near 500 °C (Peak B2), and one near 700 °C (Peak C1). Essentially, the former corresponds to Peak A1 in Figure 7a and still relates to weak hydrogen traps. The latter two peaks, namely the desorption peaks appearing at 500 °C (Peak B2) and 700 °C (Peak C1), are associated with hydrogen traps with different binding energies. However, the specific types of these hydrogen traps are difficult to determine.

Figure 7c shows the hydrogen desorption behavior during the non-uniform deformation stage. It can be observed that the relative height of the first two hydrogen desorption peaks (Peak A3 and Peak B3) decreased compared with the uniform plastic deformation stage. The height of hydrogen desorption Peak C2 at around 700 °C is still the highest. Figure 7d shows the hydrogen desorption behavior after complete tensile fracture.

Notably, the hydrogen desorption Peak A4 and Peak B4 have almost vanished, while high-temperature Peak C3 remains unchanged. This indicates that Peak A and Peak B could include contributions from other weak traps, such as grain boundaries, vacancies, and other microstructures [47]. However, this study focuses on identifying the source of the strong hydrogen traps responsible for the high-temperature Peak C.

Note that Figure 7d uses a different y-axis scale from Figure 7b,c. As the tensile strain increases, the high-temperature Peak C emerges, and its associated hydrogen content progressively rises, indicating that the number of strong hydrogen traps increases with plastic deformation. Specifically, the sample at the complete tensile fracture stage exhibits a substantially higher amount of hydrogen released from Peak C3 than samples at lower strain magnitudes. Based on microstructure characterization under different deformation stages, besides the unchanged precipitates, the difference between them lies in the fact that there are more dislocation defects around the AF (as shown in Figure 5). Thus, it could be inferred that the high-temperature Peak C3 is mainly associated with high-binding-energy hydrogen traps formed by the evolution of deformation-induced defects. A large number of dislocations proliferate and form tangled dislocations inside AF during straining, which are generally regarded as strong hydrogen traps. Shang et al. [37] also verified that tangled dislocations can act as potent hydrogen traps.

3.3. Comparison of Hydrogen Desorption from Fractured Specimens of AF-Rich and AF-Poor WMs

To directly assess whether a higher AF content leads to the formation of more strain-induced strong hydrogen traps, TDS measurements were performed on the fractured tensile specimens of both WMs (AF content 74.5% and 53.9%) after testing in 6.3 MPa H₂. Both specimens were deformed to complete fracture under identical conditions (strain rate 1×10^{-5} /s, 6.3 MPa H₂, room temperature). The TDS results are presented in Figure 8.

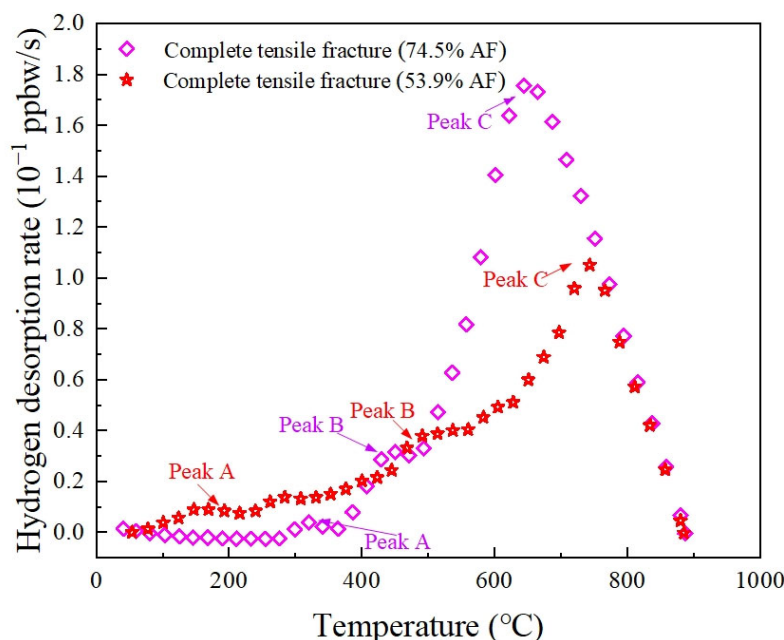


Figure 8. Comparison of TDS spectra from fractured specimens of AF-rich and AF-poor WMs tested in 6.3 MPa H₂.

For the AF-poor WM (53.9% AF), the TDS spectrum shows a prominent high-temperature desorption peak at approximately 700 °C (Peak C), accompanied by two weaker peaks at around 250 °C (Peak A) and 500 °C (Peak B). For the AF-rich WM (74.5% AF), the high-temperature Peak C is substantially more intense, with an integrated hydro-

gen content approximately 140% higher than that of the AF-poor WM (quantified by peak area integration). In contrast, the low-temperature peaks (A and B) are slightly lower in the AF-rich WM.

These results indicate that, under identical deformation and hydrogen charging conditions, the AF-rich WM traps significantly more hydrogen in strong traps (Peak C) than the AF-poor WM. Since the high-temperature peak has been associated with deformation-induced tangled dislocations within AF (Section 3.2.3), the higher intensity of Peak C in the AF-rich WM directly demonstrates that a larger volume fraction of AF generates stronger trapping sites during plastic deformation. Consequently, a greater proportion of hydrogen is immobilized in these deep traps, reducing the concentration of diffusible hydrogen that can participate in embrittlement. This provides a direct mechanistic explanation for the lower HE index of the AF-rich WM (Section 3.1).

4. Discussion

4.1. Mechanisms Underlying the Reduced HE Susceptibility in AF-Rich WM

Section 3.1 demonstrates that the WM with a higher AF content (74.5%) has a significantly lower HE index (13.4%) than the one with lower AF content (53.9%, HE index 23.5%). The detailed characterization in Section 3.2, although performed on the AF-poor WM, provides critical insights into the mechanism that can explain the beneficial role of a higher AF fraction.

In the AF-poor WM, TDS (Figure 7) revealed that plastic deformation induces a new, strong hydrogen trap (Peak C at $\sim 700^\circ\text{C}$), which becomes increasingly dominant with rising strain. Concurrently, EBSD (Figures 4 and 5) showed that AF undergoes uniform plastic deformation, grain fragmentation, and a substantial increase in GND density, especially inside AF grains. HMT (Figure 6) demonstrated that hydrogen redistributes from grain boundaries to a more uniform distribution within AF grains as deformation proceeds. These observations lead to the following interpretation: AF is not merely a static trap provider (via HAGBs and fine grains) but dynamically generates strong traps in the form of tangled dislocations during plastic deformation.

When the AF content is higher, a larger volume fraction of the microstructure contributes to this strain-induced trapping effect. Consequently, under the same applied strain, an AF-rich WM traps more hydrogen in strong, immobile states, reducing the concentration of diffusible hydrogen that can migrate to crack tips and cause embrittlement. This explains the lower HE index despite the higher overall hydrogen uptake (as suggested by the larger Peak C area in TDS). In contrast, PF deforms heterogeneously, retains hydrogen primarily at grain boundaries, and does not develop uniformly distributed strong traps, making it less effective at immobilizing hydrogen under dynamic loading.

4.2. The Dynamic Role of AF as a Strain-Induced Hydrogen Trap

Traditional studies have emphasized the static trapping ability of AF due to its high density of HAGBs and fine grain size [13–18]. The present work extends this understanding by showing that AF exhibits excellent deformation compatibility (Figure 5) and that its trapping capacity increases with plastic strain (Figure 7). This is a critical finding for service conditions where local plastic deformation is inevitable.

The emergence of the $\sim 700^\circ\text{C}$ desorption peak (Peak C) is particularly noteworthy. Its binding energy is higher than that of typical HAGBs or precipitates, which usually desorb at lower temperatures [48]. The fact that this peak intensifies with strain and persists after fracture, while lower-temperature peaks (A and B) diminish, indicates that the strong traps are directly linked to deformation-induced defects within AF—most plausibly, tangled

dislocation networks and associated vacancies or dislocation-cell boundaries. These defects are known to act as deep hydrogen traps [37].

By contrast, PF grains, being coarser and surrounded by harder AF, experience strain localization at grain boundaries (Figure 5h,i). This leads to hydrogen remaining concentrated at PF boundaries (Figure 6f), which is a classic configuration for intergranular crack initiation. Therefore, the presence of AF not only provides strong traps but also promotes a more uniform strain distribution, avoiding detrimental local hydrogen buildup.

4.3. Implications for Hydrogen-Resistant Weld Design

From an engineering perspective, this study suggests that increasing the AF content in pipeline girth welds is a viable strategy to enhance resistance to HE under dynamic loading conditions. More broadly, microstructures that can develop uniform, high-density dislocation structures during deformation (i.e., that exhibit good “deformation compatibility”) are desirable. Such microstructures act as self-tunable hydrogen traps. They are relatively less effective in the undeformed stage but become increasingly effective as deformation proceeds—exactly when hydrogen-induced cracking is most threatening.

Future weld procedure development should aim to maximize AF formation, for example, by optimizing heat input, cooling rate, and inclusion engineering to promote intragranular nucleation of AF. It is also worth exploring whether other microconstituents (e.g., fine bainite) exhibit similar strain-induced trapping behavior.

5. Conclusions

In the present work, the dynamic evolution of microstructure and hydrogen distribution in pipeline girth welds was explored, and the HE index of WMs with different AF contents was investigated. The main findings are concluded as follows:

- (1) The WM with a higher AF content (74.5%) exhibits a significantly lower HE index (13.4%) than that with a lower AF content (53.9%, HE index 23.5%), confirming that a higher AF fraction reduces HE susceptibility.
- (2) During plastic deformation, AF undergoes uniform strain distribution, grain fragmentation, and a substantial increase in dislocation density, whereas PF deforms heterogeneously with strain localized at grain boundaries.
- (3) Hydrogen distribution evolves from being concentrated at grain boundaries in the undeformed stage to a more uniform distribution inside AF grains with increasing strain, consistent with the proliferation of dislocations. In PF, hydrogen remains localized at grain boundaries.
- (4) A new, strong hydrogen trap (desorption peak at ~ 700 °C, Peak C) emerges during plastic deformation and intensifies with strain. This trap is attributed to deformation-induced defects—most likely tangled dislocations—within AF. The presence of such traps immobilizes hydrogen, reduces diffusible hydrogen, and thereby lowers embrittlement susceptibility. Thus, AF acts not only as a static trap but also as a dynamic source of strain-induced strong hydrogen traps.

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