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Hydrogen-assisted decohesion and localized plasticity in dual-phase steel

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Abstract

Hydrogen embrittlement affects high-strength ferrite/martensite dual-phase (DP) steels. The associated micromechanisms which lead to failure have not been fully clarified yet. Here we present a quantitative micromechanical analysis of the microstructural damage phenomena in a model DP steel in the presence of hydrogen. A high-resolution scanning electron microscopy-based damage quantification technique has been employed to identify strain regimes where damage nucleation and damage growth take place, both with and without hydrogen precharging. The mechanisms corresponding to these regimes have been investigated by employing post-mortem electron channeling contrast imaging and electron backscatter diffraction analyses, as well as additional in situ deformation experiments. The results reveal that damage nucleation mechanism (i.e. martensite decohesion) and the damage growth mechanisms (e.g. interface decohesion) are both promoted by hydrogen, while the crack-arresting capability of the ferrite is significantly reduced. The observations are discussed on the basis of the hydrogen-enhanced decohesion and hydrogen-enhanced localized plasticity mechanisms. We discuss corresponding microstructure design strategies for better hydrogen-related damage tolerance of DP steels.

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1. Introduction

Ferrite/martensite dual-phase (DP) steels are attractive materials for automotive applications [1–3], owing to a good combination of tensile strength (up to 1.2 GPa) and ductility (e.g. 10% or higher) [1,4–6]. This beneficial combination of mechanical properties in conjunction with a lean alloy concept is directly linked to the micromechanics of the composite-like ferrite–martensite microstructure. The high mechanical contrast inherent to such microstructures also introduces certain risk factors, namely, strong strain/stress partitioning [7–9], strain localization [10–12] and damage evolution, e.g. by martensite cracking [1,5]. While these micromechanical phenomena have been investigated in some detail, the associated hydrogen effects have not been accordingly addressed, despite reports on the susceptibility to hydrogen embrittlement (HE) of high-strength steels [13–15] and DP steels [1,16–19], and the increased susceptibility of ferrite and martensite phases to HE compared to stable austenite [20,21]. Our study aims to close this gap through a micromechanical analysis of the hydrogen-related microcracking mechanisms in DP steel.

HE in steels has been discussed mainly in terms of hydrogen-enhanced and strain-induced vacancies (HESIV)

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[22], hydrogen-enhanced decohesion (HEDE) [23–25] and hydrogen-enhanced localized plasticity (HELP) [26–29]. These embrittlement mechanisms, including their mutual interactions, are well documented in single-phase steels. For instance, the HE of fully ferritic steels was essentially ascribed to the HELP effect [30]. Lath martensitic steel, on the other hand, was reported to undergo HE associated with HELP as well as HEDE [31]. Given the more complex microstructures and local mechanics of DP steels, the connection of their HE to the HEDE, HESIV and HELP mechanisms, respectively, is expected to be more complex. In fact, indications of differences compared to single-phase materials were observed in a previous work on DP steels: unlike typical martensitic steels (which often show HE in the elastic or microplastic regimes [32–34]), HE of DP steels occurs after a significant level of plastic strain [1,16,17]. Unlike typical ferritic alloys, DP steels show significant microcracking (i.e. damage) before the final fracture even without hydrogen charging [5]. Note that this type of localized brittleness is caused by cracking of the martensite, i.e. the HE susceptibility of DP steel typically decreases by tempering [16] and increases with higher martensite fractions [17]. These observations underline the important influence of local plasticity and the damage sensitivity of the two phases on the overall HE susceptibility of DP steel. In order to elucidate these effects more clearly, we conduct here both (i) a quantitative assessment of the damage evolution and (ii) a crystallographic analysis of these microstructural phenomena in presence of hydrogen as a function of plastic strain.

Step-by-step quantitative analysis of micromechanical phenomena, such as gradual strain partitioning and damage evolution, are often not considered in conventional approaches to the evaluation of HE susceptibility of high-strength steels. Instead, certain critical macroscopic parameters (e.g. strain, external stress, stress intensity factor, hydrogen content) are typically determined when analyzing results obtained from flat [33,34] or notched [31,35,36] tensile samples. These more practical approaches are partially due to hydrogen-diffusion-related difficulties in experimentation, i.e. hydrogen can be released from the material, depending on the charging and environmental holding parameters. In order to render such analysis more quantitative and more microstructure-oriented, we investigate here the effects of hydrogen charging in conjunction with (i) electron backscatter diffraction (EBSD) based local misorientation measurements for assessing strain partitioning and (ii) a novel continuum damage mechanics-based approach to assess damage evolution,[1] both conducted through post-mortem characterization of fractured samples, without any limitation due to hydrogen desorption.

A more detailed crystallographic and local plasticity-oriented analysis of the governing damage micromechanisms, on the other hand, requires the combined use of EBSD and electron channeling contrast imaging (ECCI) [40], as was conducted in previous works on 316 stainless steel and twinning-induced plasticity steel [41–43]. Here, the post-mortem EBSD/ECCI technique is employed to retrace the hydrogen-assisted local plasticity and cracking mechanisms in DP steels. These analyses are also supported by in situ deformation experiments providing direct evidence of the observed mechanisms.

In the following, we first explain these experimental methods and the undeformed microstructures in detail. A full analysis of hydrogen-assisted damage evolution is then presented, introducing quantitative trends and the underlying micromechanisms. Finally, these results are discussed on the basis of the proposed HE theories.

2. Experimental procedure

A DP steel sheet with the chemical composition shown in Table 1 was solution treated at 1183 K for 1 h, followed by air cooling to ambient temperature. The sheet was then intercritically annealed at 1023 K for 30 min and subsequently water quenched to produce a ferrite/martensite microstructure. Tensile specimens with gauge dimensions of 1.0 mm width × 0.4 mm thickness × 5.0 mm length were machined by mechanical grinding and spark erosion.

Hydrogen charging was performed for 1 h in a 5% H_2SO_4 aqueous solution containing 3 g l^{-1} of NH_4SCN at a cathodic current density of 2 A m^{-2}. A platinum foil with dimensions of 25 mm width × 0.1 mm thickness × 25 mm length was used as the counter electrode. Since the surface area of the Pt foil in the solution was 1260 mm^2, the anodic current density was about 0.8 A m^{-2}. Since the effective diffusion coefficient of hydrogen is higher in ferrite than in martensite (i.e. due to the high density of dislocations in martensite acting as trapping sites, decreasing the diffusion speed), the diffusion coefficient of hydrogen in DP steels is considered to be higher than that in martensitic steels. The diffusion coefficient of hydrogen in martensite depends not only on the solute concentration but also on the tempering conditions [44–46]. Since the effective diffusion coefficient of hydrogen was reported to be 3.7×10^{-11} m^2 s^{-1} in an as-quenched low-alloy martensitic steel [45], the diffusivity of hydrogen in the DP steel is higher than the diffusivity of the martensitic steel. The hydrogen-affected zone can be estimated roughly by $(Dt)^{1/2}$, where D is effective diffusion coefficient of hydrogen and t is the hydrogen-charging time. Assuming that the diffusion coefficient of hydrogen in the DP steel is 3.7×10^{-11} m^2 s^{-1}, the hydrogen-affected zone just after hydrogen charging for 1 h is 0.365 mm. Since hydrogen enters from both surfaces of the specimen, it can be safely claimed that the whole specimen was affected by hydrogen charging.

Thermal desorption analysis (TDA) was carried out from room temperature to 650 K using mass spectrometry

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[1] This method relies on determining permanent damage fraction (i.e. the ratio of the damaged to the undamaged cross-sectional area [37–39]) from high-resolution SEM imaging, and the strain levels corresponding to these damage values from digital image correlation (DIC).

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