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Research Article

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Compatibility of cold drawn and heat treated austenitic stainless steel (316L) tube for green hydrogen applications

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Abstract

Austenitic stainless steels are widely used for hydrogen handling and transport due to their corrosion resistance and generally low susceptibility to hydrogen-assisted cracking. However, soft zones introduced to material during fabrication and repairs or in services remains a concern for tube-based hydrogen distribution systems. In this work, a 316L (14404) cold drawn 1/2" (12.7 mm outer diameter) tube with wall thickness 1 mm was investigated in the cold drawn state and after weld-thermal cycle simulation by joule heating (heat-treated) state. Tube sections were machined into hollow tensile specimens and tested by slow strain rate tensile (SSRT) loading at the nominal strain rates of $1\text{E-}4\text{ s}^{-1}$, $1\text{E-}5\text{ s}^{-1}$ and $1\text{E-}6\text{ s}^{-1}$ under the in-situ gaseous hydrogen and in argon as reference (150 bar, room temperature). Additional precharging conditions were performed in the ex-situ gaseous hydrogen autoclave precharging (1000 bar, 150 °C, 21 days) before SSRT.

Tensile properties (ultimate tensile strength, yield strength, elongation, and reduction of area) were quantified, and hydrogen effects were expressed using degradation indices. The SSRT results at all strain rates indicate enhancement of strength and ductility in both the cold drawn and heat-treated states. Across the investigated conditions, fracture surfaces showed predominantly ductile microvoid coalescence fracture. These findings contribute to the understanding of the compatibility of 316L tubing for hydrogen applications under the investigated conditions while highlighting the importance of processing state and hydrogen interactions.

Keywords: slow strain rate, degradation index, strengthening, deformation twinning, austenitic stainless steel, Martensite formation,

1 Introduction

Green hydrogen and its derivatives are regarded as clean, reliable and versatile energy carriers for the future. However, the supply value chain of hydrogen from production, storage, transportation, and utilization requires special and suitable materials. Amongst the common materials used are the austenitic stainless steels, due to their high corrosion resistance, toughness, ductility and relatively low hydrogen diffusion rates[1-3] as well as low susceptibility to hydrogen assisted cracking[4]. It is well documented that exposing austenitic stainless to gaseous hydrogen may cause detrimental effect and degradation of mechanical properties leading to a decreased ductility[5] and fracture toughness[6, 7] as well as increased crack growth rate[8, 9]. This influence of the mechanical properties poses a high risk for the safe transportation of hydrogen by tubings in the refueling stations. For safe hydrogen transportation, it is crucial to conduct mechanical tests in representative hydrogen environments and mimic the tubes working conditions to demonstrate the materials hydrogen degradation resistances as shown in Figure 1. It is well known that hydrogen assisted cracking (HAC) results from the synergistic interaction of a hydrogenated environment, a susceptible material and mechanical loads[10].

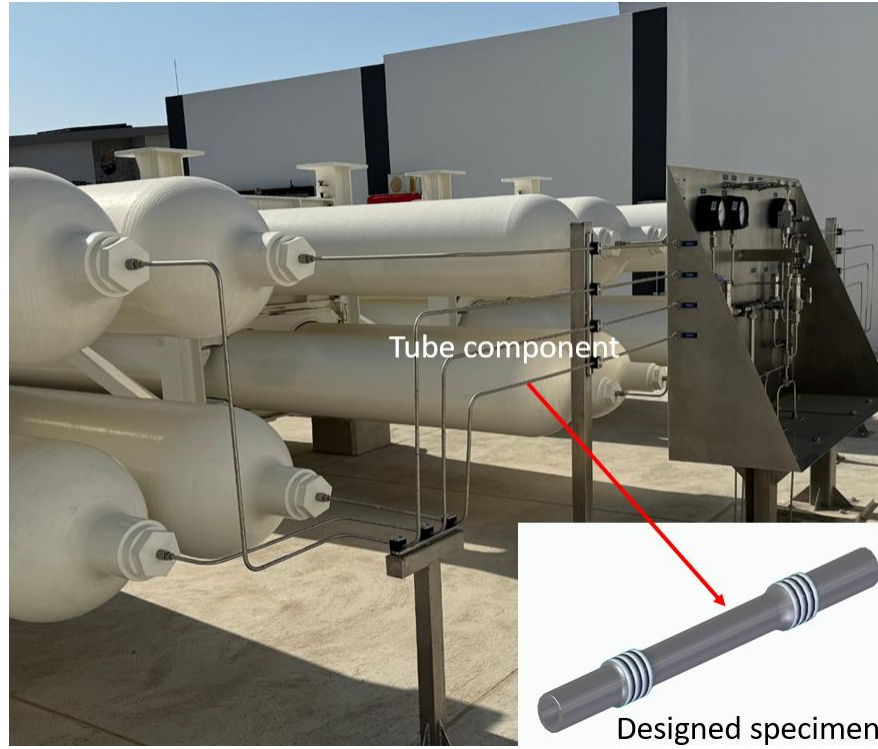


Figure 1. Hydrogen refueling station tube components conversion to specimen (not to scale)

Several well-documented HAC mechanisms proposed including hydrogen internal pressure[11-13], hydrogen enhanced localized plasticity (HELP)[11, 14, 15], hydrogen-enhanced decohesion (HEDE)[16-19], adsorption-induced dislocation emission (AIDE)[20] and hydrogen-enhanced strain-induced vacancy (HESIV)[21, 22].

Although it is well understood that diffusible hydrogen is responsible for HAC in materials, the specific mechanism with regards to austenitic stainless steels are still under debate. There are still contradictions on the undesirable effects of hydrogen on austenitic stainless steels as it may cause hardening, softening and/or both [23-26]. This study aims to investigate the hydrogen influence on the tensile properties of the 316L (1.4404) austenitic stainless steel at room temperature. The supplied tube materials were transformed into the tubular specimen in accordance with hollow specimen design. The specimens were tested in H_2 and in Argon as reference for validation with inner pressure of 150 bar, ambient temperature and nominal strain rates of $1E-4\ s^{-1}$, $1E-5\ s^{-1}$ and $1E-6\ s^{-1}$ were kept the same for all the tests meanwhile in the cold drawn and heat-treated states.

The SSRT results obtained in all tested states and conditions were analysed based on different parameters typically the reduction of area (RA) and elongation (EI) at fracture. Furthermore, analyses of the fracture morphology of the specimens were conducted using a scanning electron microscope (SEM). In addition, the benefits of employing the hollow specimen technique to evaluate the HAC susceptibility of materials in a simulated hydrogen environment are demonstrated. Therefore, the work contributes to further understanding of hydrogen-metal interactions and quantification of HAC susceptibility in order to enhance material reliability with safe deployment in advanced applications of green hydrogen industries.

2 Experimental

2.1 Material Characterisation

The tube of 1/2" (12.7mm OD) cold drawn manufactured and sourced for this study, is used as the tubing component for the gaseous hydrogen transportation within a refueling station. The chemical composition is presented in Table 1. The elemental analysis was done using optical emission spectrometry (OES) Spectrotest (method FE-30-N mode) and compared to the supplier data, both showing agreement with the standard nominal ranges. Table 2 lists the mechanical properties obtained from the supplier's certificate of the investigated material.

Table 1. Chemical composition of the investigated tube measured using OES, (all values in wt.%).

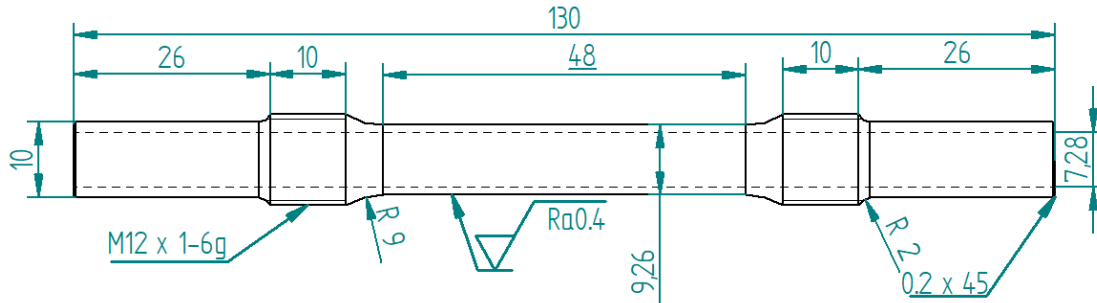
Material/Element	C	Si	Mn	P	S	Cr	Mo	N/N*	Ni	FE
Measured	0.012	0.40	1.72	0.049	0.009	16.58	1.85	0.30/0.16	11.34	Bal.
Supplier	0.012	0.39	1.66	0.034	0.009	16.61	2.02	0.051	11.11	
Nominal	≤0.03	≤0.75	≤2.00	≤0.045	≤0.030	16.00	2.00- - 3.00 18.00	≤0.100	10.00- 14.00	

Table 2. Mechanical properties of the 316L tube as per certificate provided by the supplier

316L (1.4404)	Yield strength 0.2% MPa	Tensile strength MPa	Elongation A2 (%)	Hardness HRC
Nominal value	518-758	724-965	min 20	max.30
Av. Measured	674	744	32.5	27

2.2 Specimen Preparation

The seamless tube component was converted into a tensile specimen as per design shown in Figure 2 and in adopting the hollow specimen technique[27-29]. The inner surface finishes and conditions of the specimen were kept in the same state as for the tubing real industrial operational case. All specimens were fabricated in the cold drawn (CD) state and thereafter some specimens were joule heated forming the heat-treated (HT) specimens. The CD and HT specimens were cleaned and taken for autoclave precharging in high-pressure gaseous hydrogen at 1000 bar, 150 °C, for 21 days, and are designated as CD-P and HT-P, respectively. In accordance with DIN EN ISO 3690[30], after gaseous hydrogen precharging, the specimens were stored in the liquid nitrogen (-196 °C) for hydrogen preservation for the duration not longer than three weeks before SSRT test or hydrogen content measurement. To establish the material hydrogen uptake, small coupon of tubes with length of 20 mm were cut and taken hydrogen measurement for the precharged and none-precharged in both CD and HT state. Coupons or SSRT specimen after taken out of the high-pressure autoclave charging in less than 120 s exposure time were stored in the liquid nitrogen cylinder as per the practice to prevent hydrogen egress. Before SSRT testing or hydrogen content measurement, the specimens were defrosted and cleaned in ethanol (Ultrasonic bathed) for 120 s and subsequently blown with gaseous nitrogen for drying and further residues removal for about 60 s. For the hydrogen content measurement, the coupons are weighed and the mass transferred for further measurement utilising the carrier gas hot extraction (CGHE) technique with a Galileo G8 (Bruker corporation, Billerica, USA). The CGHE coupled with mass spectrometer (MS) ESD 1000 from InProcess Instruments. The coupon (sample) is heated inside the glass tube using an infrared radiation- emitting furnace, labelled -IR07 of the G8 system. By thermally activating from ambient temperature to 800 °C, the diffusible hydrogen content present in coupon and employing a carrier gas stream of nitrogen (N₂), the hydrogen is transported to the MS detection chamber, measured and recorded accordingly.

**Figure 2.** Geometry and dimensions of the SSRT 1/2"(12.7mm OD) specimen (all values in mm)

Microstructures of the prepared CD and HT specimens are presented in Figure 3. The grain size and hardness values of all the specimens are given in Table 3. Small sample tubes were cut vertically and horizontally cut to obtain the longitudinal (LS) and cross-sectional orientation (CS), respectively. The specimens were mounted in an epoxy resin, ground, polished, and then etched with a BERAHA II color etchant. The images after etching are captured with light microscopy and finally the image data were analyzed. In addition, grain size measurement were obtained as presented in Table 3.

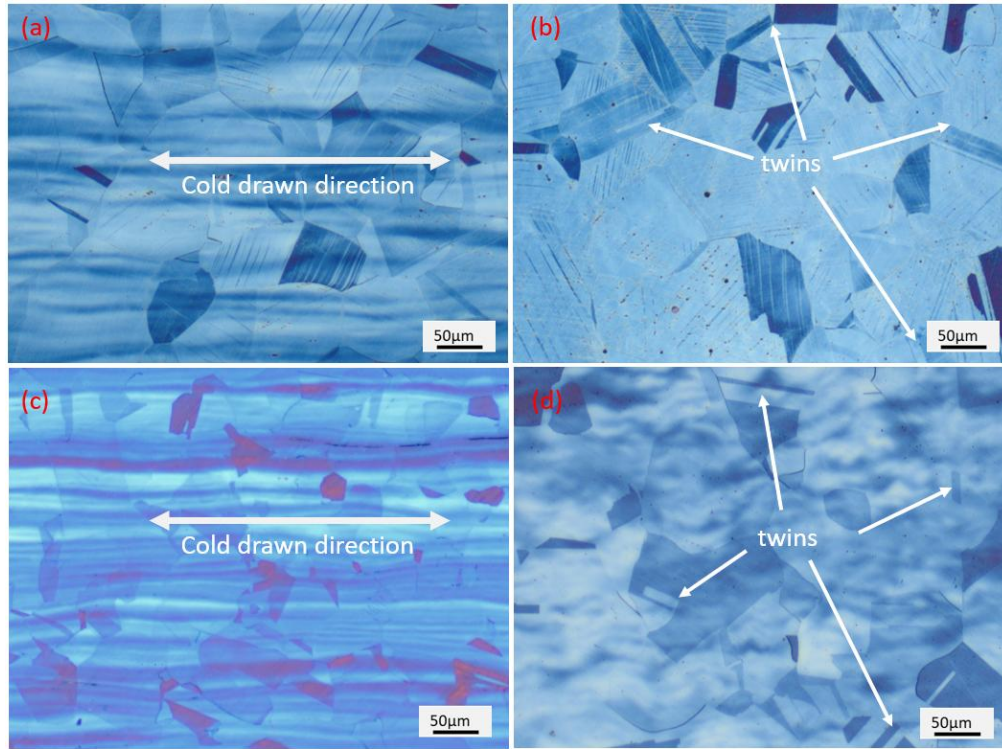


Figure 3. Microstructure of the 316L etched in BII in CD and HT state with twins and grain boundary facet:
(a) CD longitudinal, (b) CD cross sections, (c) HT longitudinal and (d) HT cross sections.

Table 3. Grain size and hardness measurement in the CD and HT state in the cross (CS) and longitudinal sections (LS).

Materials	Grain size (μm)	Area (μm^2)	Hardness (HV1)
	ASTM Norm		
CD - LS	6.7 ± 0.12	1415 ± 100	244 ± 7.4
CD - CS	7.3 ± 0.21	924.2 ± 135	241.8 ± 4.2
HT - LS	6.9 ± 0.15	1214.7	142 ± 3.29
HT - CS	6.8 ± 0.15	1343.3 ± 128	135.7 ± 3.02

2.3 Heat Treatment

Tubes for the handling and transportation of hydrogen can be mechanically assembled or welded. As a first approach to investigate the behaviour of soft zones in the cold drawn base material, an inhouse (BAM) weld simulation experimental was setup to resemble the heating and cooling cycles for the automated orbital welding of smaller tubes as indicated in Figure 4. The experimental setup offering similar working principle as the gleeble system has been utilized in joule heating to produce the HT specimens (Figure 4c). The cooling of the joule heated specimens conducted with pressurized air blowing to achieve respective heat dissipation and ensuring homogenous in the heating and cooling cycle.

Choi and Santos[31] reported that austenitic stainless steels may form a coarse grained HAZ at temperatures ranging from 1000 °C to 1350 °C. Higher heat treatment temperature (≥ 1100 °C) induces twins and grain boundary faceting-roughening transitions. The thermal effects arise from the rapid temperature increase as current passes

through the metal. Yang et al[32] reported that joule heating can drive microstructural evolution and induces thermal softening. Meanwhile, temperatures of above 1100 °C is known to cause wetting transition in 316L, which could be detrimental to the setup and experimental work in general. In addition, due to the geometric factor and test conditions, the joule heating was set at 1100 °C with the holding time of 1s of the cooling time 800 °C to 500 °C ($t_{8/5}$). High temperatures could lead to a bending effect, moreover due to the cooling with pressuring air bowing may affect the specimen straightness. This may influence the general SSRT testing and results due to possible misalignment. Joule heating clearly results in grain growth (coarsening), leading to loss of strength and hardness. Hardness declines significantly by 45% from the CD (244 HV) to HT state (135 HV), respectively, as presented in Table 3 as per measured results obtained.

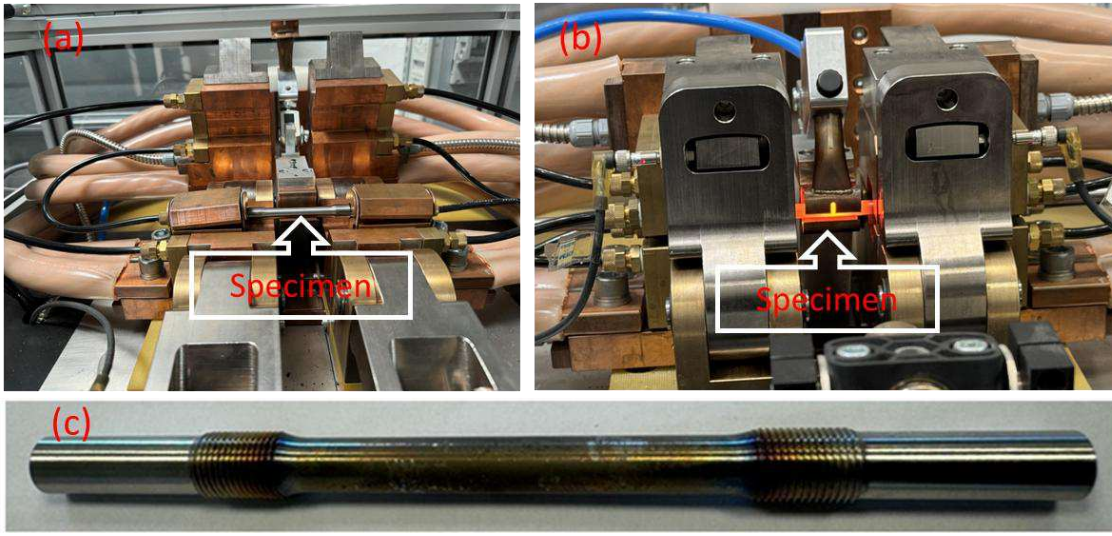


Figure 4. Heat treatment (Welding simulation) (a) Before, (b) during and (c) specimen after joule heating (not to scale)

2.4 SSRT

Slow strain rate testing (SSRT) developed by Parkins[33], is often recommended for quantifying the effects of gaseous hydrogen fracture as the slow strain rate allows for the significant diffusive transport of hydrogen in austenitic steels in comparison to the general tensile testing. The specimens for the investigation were designed and fabricated in the BAM machining workshop from the supplied 6m long 1/2" and 1mm thickness of the gauge length section. The tensile specimen tested in the original CD and HT states were examined in the in-situ and ex-situ gaseous hydrogen charging of 200 bar at room temperature and 1000 bar at 150 °C for 21 days. The tests are performed in strain rate sets of $1\text{E-}4\text{ s}^{-1}$, $1\text{E-}5\text{ s}^{-1}$ and $1\text{E-}6\text{ s}^{-1}$ adopting the hollow specimen technique[34, 35]. The SSRT tests were performed with a Zwickroell Kappa 150 DS machine in accordance with the DIN EN ISO 6892-1:2020-06[36]. Before testing, the samples were rinsed in acetone with purity of 99.8 %, the hollow was mechanically cleaned and afterwards rinsed in ethanol with a purity of 99.2 % and dried in gas nitrogen flow. The samples were filled with gaseous hydrogen in a purity of 5.0 (99.999 %) and purged cycles 10 times by filling the specimen with 150 bar, then released to 10 bar, thereafter the final refill of 150 bar pressure during the testing. The overall duration of the purging and refilling process was about 5 to 10 min and thereafter the mechanical loading was started immediately after the final refilling. Nitrogen gas was used in the initial purge before hydrogen refill. All reference tests were conducted using argon of 5.0 (99.999 %) purity, 150 bar, at room temperature. All the tests, strain was measured using a clip-on Epsilon 3542 axial extensometer with the gauge length of 35 mm. The preload of 300 N was applied just prior to the starting of all the tensile test to eliminate the default machine error when using the extensometer. During SSRT tests all the specimens were filled with either hydrogen or argon as a reference. The CD specimens that were filled with argon or hydrogen are designated as CD-Ar and CD-H₂, respectively. Similarly, the CD precharge specimens that were filled with argon or hydrogen are designated as CD-P-Ar and CD-P-H₂, respectively. Additionally, the HT specimens that were filled with argon or hydrogen are

designated as HT-Ar (black) and HT-H2 (blue), respectively. Similarly, the HT precharge specimens that were filled with argon or hydrogen are designated as HT-P-Ar (green) and HT-P-H2 (red), respectively. All the tests in hydrogen environment were repeated two to three times in order to ensure the reproducibility and reliability of the experimental results, while in argon environment as reference were tested only once.

2.5 Hydrogen Degradation Measurement

Hydrogen susceptibility can be quantified by comparing changes in values of measurement in the original state before exposure to hydrogen and after exposure or alternatively comparison of the test results in an inert environment (reference) to the test results in the hydrogen environment based on ductility (reduction of area and elongation) using Equations (2) and (3) or strength (UTS) using Equation (4), respectively. Standards such as ASTM G129[37] and ASTM G 142-22[38] all indicates UTS and elongation (EL) at fracture as potential parameters to measure hydrogen degradation, which can be obtained from the stress-strain curves.

After SSRT testing, the fracture surfaces were analysed to determine the size of the cross-sectional areas using a digital light microscopy (Keyence VHX-5000, Osaka, Japan). The measured values were used to calculate the reduction of area (RA) using equation (1)[39].

$$RA[\%] = \frac{A_o - A_f}{A_o} * 100 \quad (1)$$

Where A_o is the original cross-section of the specimen before testing and A_f is the fractured cross-section after testing (fracture). Moreover, the averaged reduction of areas was used to determine the hydrogen degradation index (HDI), as shown in equation (2)

$$HDI_{RA}[\%] = \frac{RA_{ref} - RA_H}{RA_{ref}} * 100 \quad (2)$$

Where RA_{ref} represent the reduction of area tested in reference and RA_H represent the reduction of area for the specimen tested in hydrogen. Furthermore, relative reduction of area values are obtained as ratio of the reduction of area between the specimen tested in argon to hydrogen environment as in-situ or ex-situ charged, tentatively. The close the ratio ($R \approx 1$ indicates no degradation effect) is close to 1 the higher the resistance the material to the HAC influence.

$$HDI_{El}[\%] = \frac{El_{ref} - El_H}{El_{ref}} * 100 \quad (3)$$

Where El_{ref} refers to the elongation at fracture measured in reference test and El_H represent test in hydrogen environment.

$$HDI_{UTS}[\%] = \frac{UTS_{ref} - UTS_H}{UTS_{ref}} * 100 \quad (4)$$

Where UTS_{ref} refers to the UTS measured in reference test and UTS_H represent test in hydrogen environment.

3 Results and Discussion

3.1 SSRT

Hydrogen content measurements were conducted and recorded as 1.9 ± 0.66 and 5.5 ± 0.14 ; 78 ± 4.01 and 82 ± 10.77 ppm of the HT and CD non-precharged and precharged, respectively. This record proven that the material regardless the state does take-up comparatively more hydrogen during the gaseous high pressure precharging. This helps to disregard the tendency of austenitic stainless steel developing a chromium oxide protective layer that may inhibit hydrogen diffusion in the material. The influence of hydrogen on the mechanical properties has been

assessed in terms of degradation index and elongation loss by comparing the test results in hydrogen and Argon at the strain rates of $1\text{E-}4\text{ s}^{-1}$, $1\text{E-}5\text{ s}^{-1}$ and $1\text{E-}6\text{ s}^{-1}$. Tensile properties using the stress-strain curves are presented in Figure 5, shows the results of all the tested specimens. The hydrogen precharged specimens reveals an increase in the UTS and elongation (ductility) as shown in Figure 5a, 5c and 5e in the CD and Figure 5b, 5d and 5f in the HT state in all the measured strain rates, respectively. The UTS measured in the CD regardless strain rates, non-precharged and precharged are 786.58 ± 22.72 , 769.5 ± 12.32 , and 803.67 ± 16.82 , meanwhile in the HT are 592 ± 20.53 , 577.17 ± 10.8 and 606.83 ± 16.82 , respectively. Furthermore, UTS records per strain rate starting with $1\text{E-}4\text{ s}^{-1}$ in the sequence as total measured per state inclusion of all conditions, non-precharged and precharged in the CD are 798.75 ± 23.41 , 779.5 ± 12.02 and 818 ± 4.24 , concurrently, in HT state are 603 ± 17.76 , 588.5 ± 9.19 and 617.5 ± 4.95 , sequentially. The UTS with $1\text{E-}5\text{ s}^{-1}$ in the CD (all conditions), non-precharged and precharged record are 791.5 ± 21.89 , 773 ± 4.24 and 810 ± 7.07 , meanwhile for HT are 595.5 ± 25.65 , 573.5 ± 3.54 and 617.5 ± 4.95 , respectively. Finally, the UTS in the $1\text{E-}6\text{ s}^{-1}$ in the CD (inclusive of all condition), non-precharged and precharged recorded as 769.5 ± 15.7 , 756 ± 2.8 and 783 ± 1.41 and for the HT are 577.5 ± 10.66 , 569.5 ± 9.19 and 585.5 ± 0.71 , The results reveal a distinct increase of less than 7 % in the UTS for the precharged compared to the non-precharged both in the CD and HT state at all the strain rates.

Additionally, more results are indicated in Figure 6 (a) UTS, (b) YS, (c) RA, and (d) time to failure for further comparison of the tensile properties. The CD shows a predominantly higher UTS and YS compared to HT state as presented in Figure 6 (a) and (b), respectively. Moreover, the UTS and YS in the precharged environment is slightly higher compared to non-precharged environment in both the CD and HT, in all strain rates. The YS increases by less than 23% in the precharge compared to non-precharge across all states and strain rates. Contrary, a decrease of less than 8% in the YS is also observed. In addition, there is an increase of less than 5% YS of the non-precharged in-situ specimens tested in hydrogen and argon (reference) condition.

Figure 6c, indicated a predominantly higher RA in the HT compared to CD state and higher in the non-precharged compared to the precharged in all the tested strain rates. The measured relative reduction of area ratio ranges from 0.84-1.05, indicating a higher resistance to degradation of the tested materials under all conditions. There is no clear trend or influence in RA regarding comparison between the different strain rates. Figure 6d, shows the test (duration) time to failure which indicated less time as the strain rates increase, meanwhile, longer duration of test with the HT compared to CD and relatively more time in the precharged compared to non-precharged conditions.

Hydrogen degradation indices are presented in Figure 7 (HDI_{RA}), Figure 8 (HDI_{EL}) and Figure 9 (HDI_{UTS}) as obtained from the Equations (1)-(4). All the indices are calculated and comparable in the six test condition variations as Ar vs H₂, Ar vs H₂-P-Ar, Ar vs H₂-P-H₂, H₂ vs H₂-P-Ar, H₂ vs H₂-P-H₂ and H₂-P-Ar vs H₂-P-H₂, respectively, tested in all both states and strain rates. Figure 7, indicates a higher RA degradation index in CD compared to HT states in all tested conditions and strain rates. Furthermore, a predominantly positive values in the RA index is deemed to indicated that the material has undergone hydrogen degradation, meanwhile, negative values indicate the possibility of higher ductility (softening). Degradation index of elongation in the CD is higher compared to the HT state, moreover, lots of recorded values are negative which may indicate the loss ductility. Furthermore, some positive values are present indicating the possibility of increase in ductility (softening) as shown Figure 8. The hydrogen degradation index of the UTS, results show a positive increase of less than 3% in both the CD and HT state, in all the test conditions and strain rates. Predominantly, results of the UTS Index shows the negative values in both CD and HT states which indicates the possibility of strengthening of material.

Hydrogen may trigger a strain-induced martensitic transformation as a result of combination of effects from stress and hydrogen which cause a phase transformation of the austenitic (FCC) to martensitic (BCC) structure which is more brittle. 316L series of the metastable austenitic steels undergoes an unstable thermodynamically transformation when under stress and at low temperature. The formation of martensite initially relieves the internal stresses and finally hardens the material, leading to the increases in both ductility and strength as observed in the results presented in Figure 5. This leads to an assumption about the dual role of hydrogen in promoting both strengthening and softening in agreement with the findings of Birnbaum and Sofronis [15, 40]. Moreover,

contradictory experimental observations of strengthening and softening exist even in same type of material and typically noted in austenitic stainless steels[41].

Hydrogen induced softening or strengthening is dependent on a combination of complex factors such as microstructure, hydrogen content, diffusivity and experimental loading conditions. According to Ogawa et al[42], the increase in strength (UTS) due to hydrogen precharging is deemed to be a mere consequence of solid-solution strengthening. Meanwhile, the enhancement in ductility is being ascribed to an increased density of mechanical twins. SSRT results indicate that the tested material has a high resistance to hydrogen assisted cracking regardless the testing done under severe conditions to worst case scenario of high temperature and high-pressure gaseous charging. The observation both strengthening and softening effect are confirmed in this experimental, however cannot quantify which pertaining mechanisms contributes to the observed effect. From literature [31, 41, 43-47], there are interplay and synergical influences from different microstructural processes such as dislocation accumulation and interaction, mechanical twinning, phase change, solid solution that may all contribute to strengthening. Moreover, softening may arise from dislocations rearrangement and annihilation, grain recovery and recrystallization, thermal or adiabatic at high strain rates.

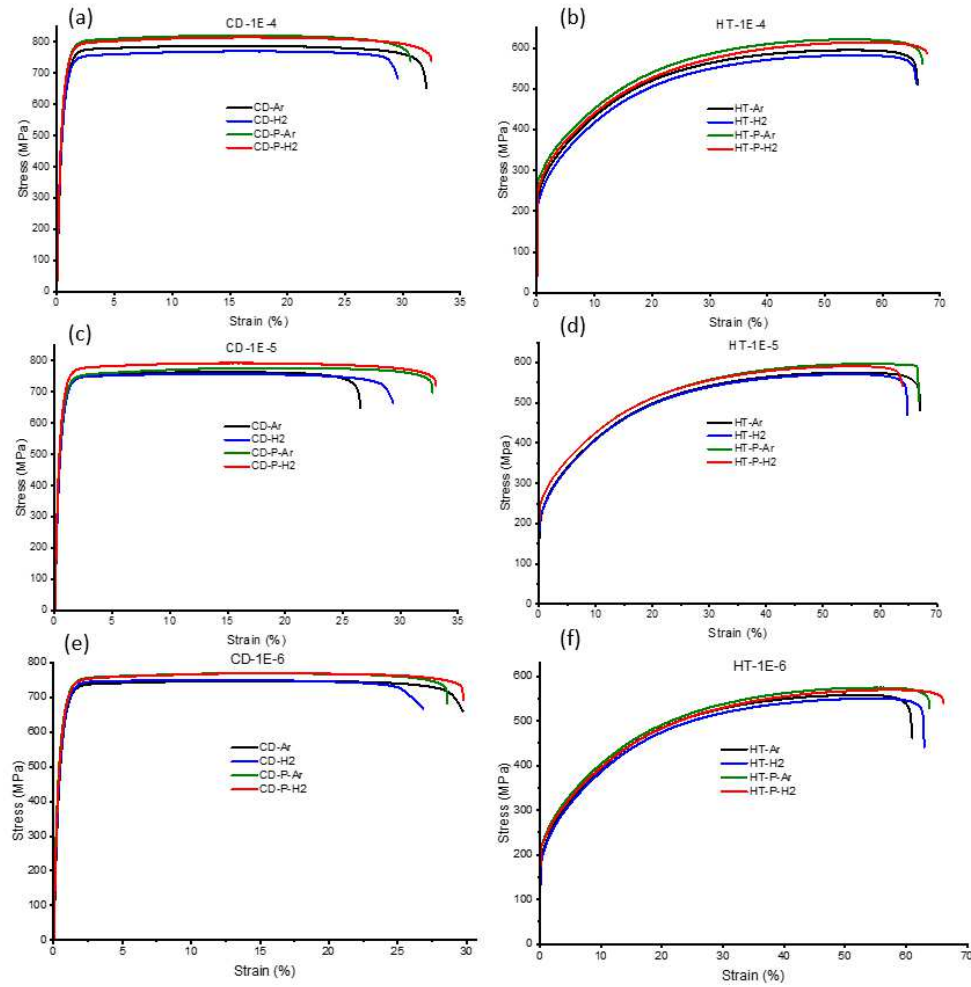


Figure 5. Stress-Strain relationship for the tested specimens in the two states and three different strain rates (a) CD-1E-4, (b) HT-1E-4, (c) CD-1E-5, (d) HT-1E-5, (e) CD-1E-6 and (f) HT-1E-6

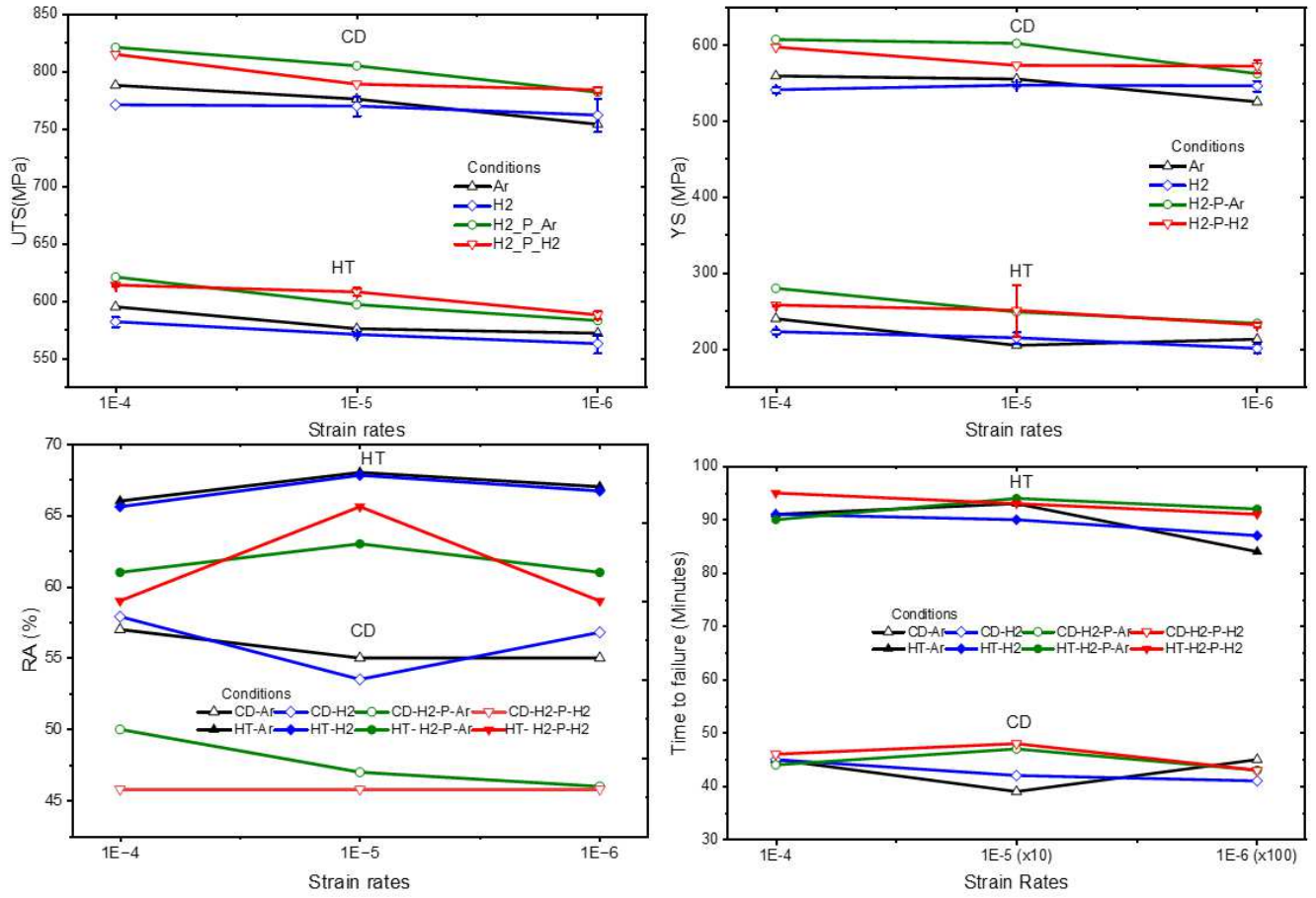


Figure 6. Tensile properties at difference strain rates and test conditions of the CD and HT (a) UTS, (b) YS, (c) RA, (d) Time to failure

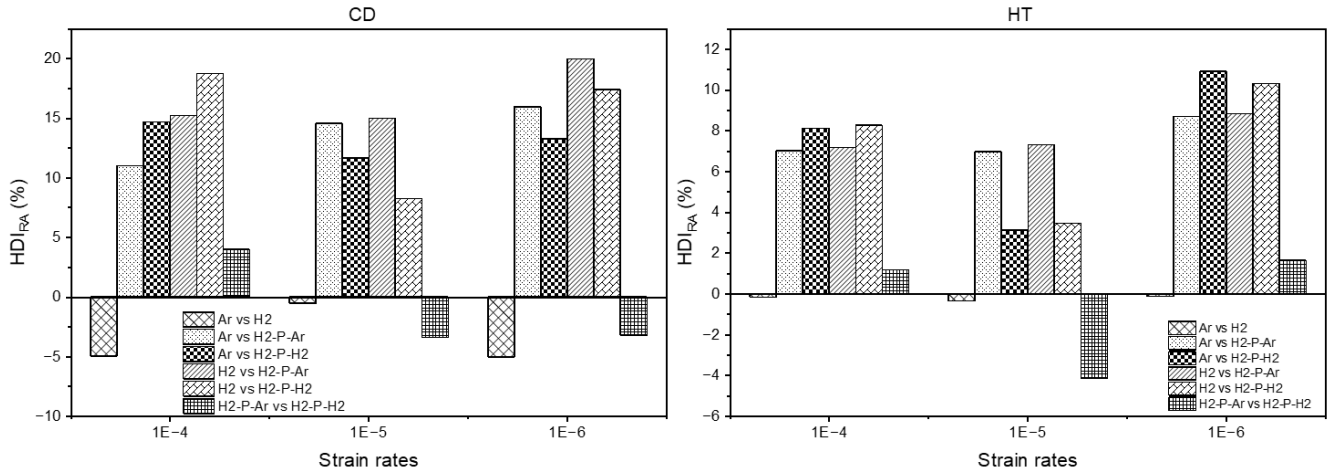


Figure 7. Degradation index of the RA in the CD and HT state, comparing the different strain rates and test conditions

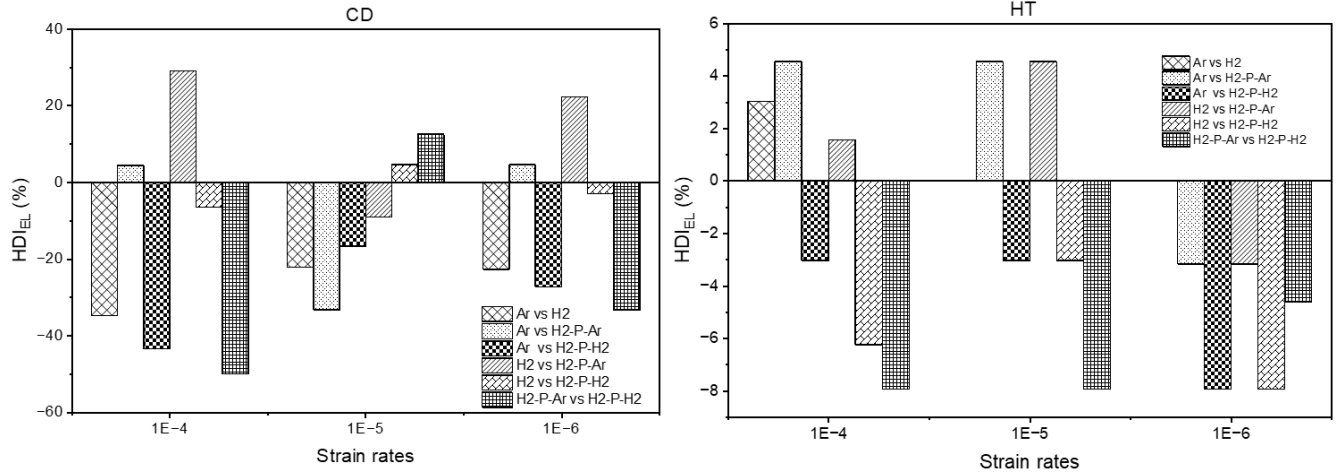


Figure 8. Degradation index of the EL in the CD and HT state, comparing the different strain rates and test conditions

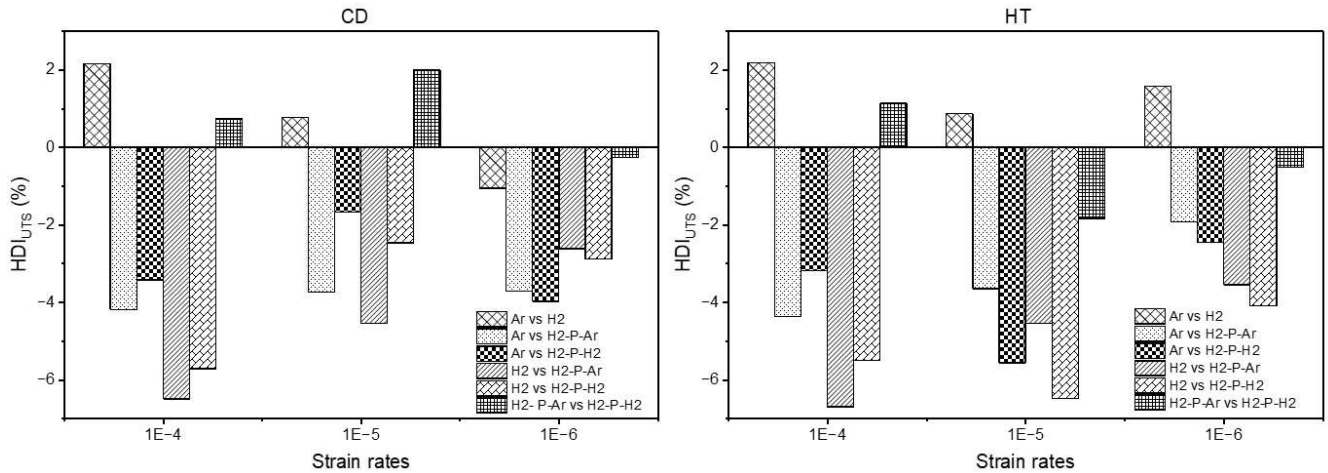


Figure 9. Degradation index of the UTS in the CD and HT state, comparing the different strain rates and test conditions

3.2 Fractographic Analysis

The fracture surfaces of sampled specimens were characterized using SEM to identify the dominant fracture features or mechanism as presented in Figure 10. The CD and HT specimens tested in different charging conditions as hydrogen (ex-situ and in-situ charged) and argon for reference (in-situ) in the three strain rates were evaluated. Random sampled figures in the CD and HT of the precharged and non-precharged at all the tested strain rates with varied condition were observed. CD state: Figure 10 (a) Hydrogen in-situ $1\text{E-}4\text{ s}^{-1}$ no high pressure charged, (b) Hydrogen precharged $1\text{E-}5\text{ s}^{-1}$, with in-situ hydrogen and (c) Hydrogen precharged $1\text{E-}6\text{ s}^{-1}$, with in-situ Argon. Meanwhile the HT state: Figure 10 (d) Argon in-situ $1\text{E-}6\text{ s}^{-1}$ no high pressure charged, (e) Hydrogen precharged $1\text{E-}5\text{ s}^{-1}$, with in-situ Argon and (f) Hydrogen precharged $1\text{E-}5\text{ s}^{-1}$, with in-situ hydrogen. Generally, fractographic examination can reveal the hydrogen degraded properties provided where cracking proceeds by all microstructural modes, such as ductile fracture, quasi-cleavage, transgranular cleavage and brittle intergranular fracture. It is observed that all the fracture surfaces belong to the fracture feature of dimple morphology. However, the surfaces appear to be non-uniformly ductile in the precharged and non-precharged environment in all states and across all the tested strain rates.

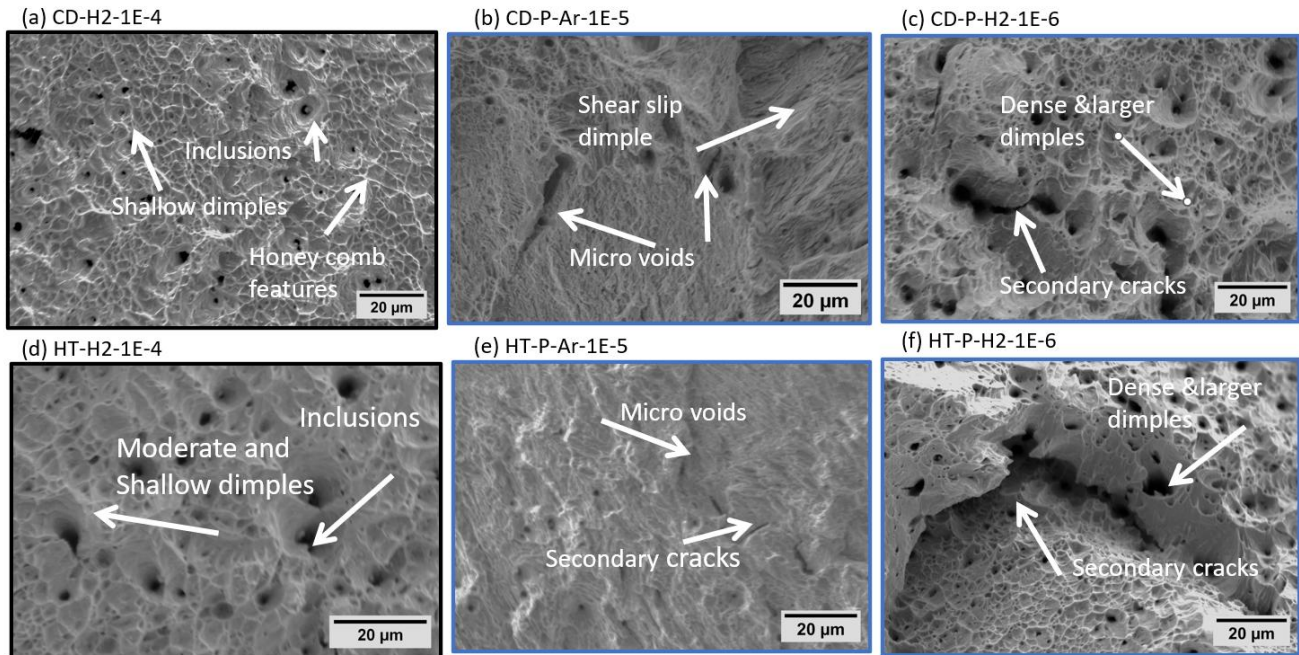


Figure 10. SEM morphology of fracture surfaces after SSRT are different strain rates and states

4 Conclusions

The hollow specimen hydrogen testing technique was adopted for the evaluation of the tensile properties of the 1/2" (12.7mm) tubing for green hydrogen application (component to specimen conversation). The specimens in the CD and HT states were slow strain tensile tested at three different strain rates in the hydrogen and reference gas environment for the assessment of Hydrogen compatibility and susceptibility of the tube component for H₂ transport purposes in the refueling station. SSRT shows no significant effect of hydrogen degradation was observed. Meanwhile the pronounced increase ductility is attributed to an increased density of mechanical twins, whereby dynamic hydrogen dislocation interactions which accelerated the twin nucleation which reduce dislocation motion in the microstructure. The conclusion can be summarized as follows:

1. The UTS and YS decreases proportionally to the strain rates regardless the states as cold drawn and heat treated.
2. The UTS and YS is predominantly higher in the hydrogen precharged in all tested strain rates and states compared to the non-precharged.
3. Strains at UTS is lower in the cold drawn compared to the heat-treated state independent of the testing strain rates with no significance difference between the precharged and non-precharged specimens.
4. Reduction area is higher in the heat treated (60-67%) than the cold drawn (46-60%) state and lower in the precharged compared 46-50% CD; (60-65%) HT to the non-precharged (54-60 %) CD; 65-68% HT at all strain rates respectively.
5. There is no significance influence of hydrogen degradation owing to the Index of below 3%.

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Declarations

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