Review

Review and Assessment of the Effect of Hydrogen Gas Pressure on the Embrittlement of Steels in Gaseous Hydrogen Environment

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Abstract: Hydrogen gas pressure is an important test parameter when considering materials for high-pressure hydrogen applications. A large set of data on the effect of hydrogen gas pressure on mechanical properties in gaseous hydrogen experiments was reviewed. The data were analyzed by converting pressures into fugacities (f) and by fitting the data using an $f^{|n|}$ power law. For 95% of the data sets, $|n|$ was smaller than 0.37, which was discussed in the context of (i) rate-limiting steps in the hydrogen reaction chain and (ii) statistical aspects. This analysis might contribute to defining the appropriate test fugacities (pressures) to qualify materials for gaseous hydrogen applications.

Keywords: hydrogen embrittlement; gaseous hydrogen; pressure effect; steel

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1. Introduction

Mobility based on hydrogen-powered vehicles such as fuel cell vehicles is a promising way to reduce greenhouse emissions. For high market penetration, the prognosis is that the costs of a hydrogen infrastructure are significantly lower compared to a battery-charging infrastructure [1]. However, there are special issues associated with the realization of a hydrogen infrastructure. One issue is the susceptibility of most structural materials, especially steels, to hydrogen embrittlement. Despite this issue, the overall goal is to realize safe and affordable infrastructure and mobility.

For automotives as well as for many infrastructure applications, the design must be safe, affordable and lightweight. The qualification of materials for use in hydrogen applications is specified in relevant standards. It is generally accepted that a material's qualification for gaseous hydrogen applications should be performed under conditions reflecting the final use of the material, i.e., in gaseous hydrogen atmospheres. Material testing in gaseous hydrogen atmospheres requires very specialized equipment [2,3] as well as strong safety protocols. Altogether, this results in high technological efforts, reflected by the costs being 15 to 150 times more expensive compared to tests in ambient air, depending on the test method (tensile test, crack growth test, etc.) [4]. Generating and maintaining a high hydrogen gas pressure throughout the test duration remains a technical challenge, especially at sub-ambient temperatures. The primary motivation of this study was to review the experimental data investigating the influence of external gas pressure on the degree of hydrogen effects in materials. The data were analyzed to quantify the effect of high hydrogen gas pressures of up to 100 MPa on the mechanical properties of steels and to propose a rationale for the definition of test pressures for high-pressure hydrogen applications.

2. Experimental and Analytical Details

This study reviews the experimental results of the influence of external high-pressure hydrogen on the various mechanical properties of steels. This study only reviews the

results obtained in gaseous hydrogen environments at room temperature, with a focus on hydrogen gas pressures of up to at least 70 MPa (Table 1). Such experiments were typically performed with apparatuses similar to those described in [2,3]. Studies investigating the influence of hydrogen gas pressure at temperatures other than room temperature were not found in the existing literature. Only results obtained by the hydrogen pre-charging of steels (gaseous or electrochemical) simulating external high hydrogen gas pressures were used to interpret the results.

Since the effect of hydrogen on the mechanical properties of steels typically increases with increasing hydrogen gas pressure (p), resulting in increased hydrogen concentration (c) [5–7], the experimental data must be fitted using an equation reflecting the rate-limiting steps in environmental hydrogen experiments. Such rate-limiting steps are [8]:

1. Transport of hydrogen to the crack tip, proportional to p ;
2. Physical adsorption, proportional to p ;
3. Dissociative chemical adsorption, proportional to $p^{0.5}$;
4. Hydrogen absorption, Sievert's law, proportional to $p^{0.5}$;
5. Transport of hydrogen to regions of tensile stress, not dependent on p ;
6. Hydrogen–material interactions, not dependent on p .

At high pressures (and low temperatures), the behavior of a gas deviates from that of an ideal gas. The appropriate thermodynamic parameter is fugacity (f), which is correlated to the pressure according to

$$f = pe^{\frac{pb}{RT}} \quad (1)$$

with b = constant, R = universal gas constant, and T = temperature [9]. The correlation between pressure and fugacity is shown in Figure 1a. At room temperature, $f \approx p$ for pressures lower than 1 MPa, which is known as Sievert's law. For higher pressures, the non-ideal gas behavior must be considered especially in the power laws of rate limiting steps 3 and 4. Using Sievert's law as an example, $p^{0.5}$ translates to $f^{0.5}$ [9].

Figure 1. Correlation between pressure and fugacity at room temperature calculated by Equation (1).

In the following, the data were analyzed in terms of fugacity instead of pressure. Based on this analysis, the experimental data were fitted using the following equation:

$$HEI \sim mf^n \quad (2)$$

where HEI = any hydrogen embrittlement index, m = the factor and n = the exponent. Typical hydrogen embrittlement indices use the ratio of the mechanical property measured in H_2 and in air, in percent. In this review, the relative fracture toughness (K_{H_2}/K_{air}), relative crack growth rate ($da/dN_{H_2}/da/dN_{air}$), relative reduction of area ($RRA = RA_{H_2}/RA_{air}$) of the tensile specimens and the relative number of cycles to the failure ($N_{f_{H_2}}/N_{f_{air}}$) of fatigue life tests were calculated based on the published experimental data. In most references, the mechanical property in the control atmosphere (in most cases, air) was not measured as a function of pressure. In other words, the value measured in ambient air was used to calculate the embrittlement index for all hydrogen fugacities. Using such indices, the degree of embrittlement increases as the index decreases, except for the crack growth tests, because crack growth is accelerated in hydrogen compared to air. The experimental data were fitted using the method of least squares. The most important experimental details of the results reviewed here are summarized in Table 1, including the fit exponent (n) from Equation (2), as well as the coefficient of determination (R^2). Further details can be found in the respective references.

3. Results

Figure 2 shows the quasi-static properties, i.e., the relative reduction of area (RRA) of the tensile specimens for heat-treatable steels (Figure 2a,b), carbon steels (Figure 2c) and austenitic stainless steels (Figure 2d,e), as well as the relative fracture toughness (relative K) of heat-treatable and low alloyed steels (Figure 2f,g) as a function of hydrogen fugacity. Microstructures, tensile properties in air, fit parameters according to Equation (2), as well as the coefficient of determination (R^2), are summarized in Table 1. All experimental data were fitted using Equation (2) with a reasonably high R^2 value greater than approximately 0.6. For most of the data, the R^2 value was greater than 0.8. The analysis of the absolute value of the fit exponent reveals a range of $|n|$ between approximately 0.05 and 0.3.

Figure 2. Cont.

Figure 2. Results of the quasi-static tensile and fracture toughness tests as a function of hydrogen fugacity. Relative reduction of area (RRA) in the tensile tests for (a,b) heat-treatable steels, (c) carbon steels, (d,e) austenitic stainless steels. Relative fracture toughness of (f) heat-treatable steels, (g) low alloyed pipeline steels. References, test details and fit parameters are given in Table 1.

Figure 3 summarizes the cyclic properties as a function of hydrogen fugacity, i.e., the relative crack growth rates (relative da/dN) for carbon steels (Figure 3a), heat-treatable steels (Figure 3b) and pure iron (Figure 3c), as well as the relative number of cycles to the failure of heat-treatable steels (Figure 3d). It should be noted here that the trend for the X80 pipeline steel was extrapolated (Figure 3d). Extrapolation of trends is generally associated with a high risk. However, the relative N_f at a fugacity of 8 MPa was already at a very low value of 0.1 and a significant further reduction in N_f with increasing fugacity was not expected. This allowed a meaningful extrapolation of the trend for this specific data set. Cyclic data systematically studying the influence of pressure for austenitic stainless steels were not found in the existing literature, but results in [10] suggest a negligible influence between 10 and 103 MPa. All data in Figure 3 were fitted using Equation (2) with a reasonably high R^2 value greater than 0.55. For most of the data, the R^2 value was greater than 0.9. The two different results for carbon steel SM490B were remarkable and even surprising (Figure 3a). Although the reported mechanical properties of the steels were very similar and the test conditions were identical (Table 1), significantly different crack growth rates were reported. A reason for this discrepancy could not be identified based on the available information in the respective references. Most relative da/dN values were calculated at $\Delta K \approx 25 \text{ MPa m}^{0.5}$. For steel HY-100, the data were acquired at $\Delta K = 55 \text{ MPa m}^{0.5}$ (Figure 3b). Quasi-static fracture toughness values between 30 and 80 $\text{MPa m}^{0.5}$ measured in high-pressure hydrogen were reported in [11] for steels with similar tensile properties. In other words, the cyclic stress intensity range of $\Delta K = 55 \text{ MPa m}^{0.5}$ might be close to the fracture toughness of HY-100 and the increasing relative da/dN values for fugacities higher than 100 MPa might be explained by approaching the transition region to unstable crack growth at such high fugacities. For the cyclic data sets, the analysis of the absolute value of the fit exponent revealed a range of $|n|$ between approximately 0.08 and 0.4. (Table 1). Based on this analysis, the $|n|$ values for cyclic tests and quasi-static tests are in the same range (Table 1).

Figure 3. Cont.

Figure 3. Results of the cyclic crack growth tests (da/dN) and fatigue life tests measuring the number of cycles to failure (N_f) as a function of hydrogen test pressure. Relative da/dN for (a) carbon steels, (b) heat-treatable steels and (c) pure iron. (d) Relative number of cycles to the failure of heat-treatable steels. References, test details and fit parameters are given in Table 1.

Table 1. Steel designations, similar grades for other regions and mechanical properties of the steels reviewed in this study. Normal letters refer to data explicitly given in the respective reference. (*) refers to data which were estimated from graphs or heat treatment conditions given in the respective references. F = ferrite, P = pearlite, TM = tempered martensite, B = bainite, AF = acicular ferrite. N_f = number of cycles to failure in Wöhler-type fatigue life tests, K = fracture toughness, da/dN = crack growth rate, na = not annotated.

Ref.	Steel Designation		Micro-Structure *	Tensile Properties in Air				Mechanical Property Tested in H ₂	Fit Parameter		
	As Given in Ref.	Equiv. Grades *		YS, MPa	UTS, MPa	E _f , %	RA, %		m	n	R ²
[12]	Ck22 normalized	1.1151, AISI 1022, S22C	F (+P)	350 *	531	32	64	tensile RA	76,236	−0.0920	0.8435
[13]	S15C	1.1141, AISI 1015	F (+P)	320 *	490 *	32 *	68 *	tensile RA			
[14]	S15C	1.1141, AISI 1015	F (+P)	na	na	na	Na	tensile RA			
[15]	S25C	1.1158, AISI 1025	F (+P)	334	547	31	72	tensile RA			
[16]	S25C	1.1158, AISI 1025	F (+P)	289	463	na	63	tensile RA			
[17]	ASME SA-106 Gr B	1.0352, SFVC 1	F + P	297	518	28	70	tensile RA	81,863	−0.1600	0.5871
[18]	ASME SA-515 Gr 70	1.0481, SGV410	F + P	338	504	na	66	tensile RA			
[19,20]	SCM440	1.7225, AISI4140	F + P	350 *	620 *	23 *	49 *	tensile RA	110,000	−0.0940	0.6595
[21]	ASTM A-302 Gr B	1.5414, SBV 1 B	TM	550 *	813	28	70	tensile RA	82,568	−0.1360	0.7540
[19,20]	SCM440	1.7225, AISI4140	TM	550 *	980*	15 *	35 *	tensile RA			
[22,23]	ASTM A533-B	1.6311, SOV 2 B	TM	703	786	20	69	tensile RA	97,583	−0.1460	0.9530
[24]	ASTM A-517-64 Gr F (T-1)	1.8881, SEV 245	TM	751	813	25	66	tensile RA			
[17]	ASME SA-372 Gr J	1.7225, SCM440	TM	725 *	883	20	60	tensile RA	94,268	−0.3120	0.7636
[25]	SCM435	1.7220, AISI 4130	TM	770 *	873 *	14 *	56 *	tensile RA			
[14]	SCM440	1.7225, SA-372 Gr J, AISI 4140	TM	Na	na	Na	na	tensile RA			
[26,27]	SUS316	1.4404, AISI316	A	Na	na	Na	na	tensile RA	88,440	−0.1250	0.9784
[27]	SUS304	1.4301, AISI304	A	Na	na	na	na	tensile RA	62,878	−0.1510	0.9459
[28]	AISI 304L	1.4306, SUS 304L	A	250 *	550 *	60 *	na	tensile RA	70,536	−0.1240	0.8557
[29]	1.4301	AISI304, SUS304	A	211	589	90	85	tensile RA	50,123	−0.0850	0.9459
[29]	1.4404	AISI316, SUS316	A	201	575	81	84	tensile RA	91,568	−0.1330	0.8383
[11]	ASTM A372	1.7225, SCM440	TM	586	813	19	49	K	72,392	−0.0970	0.5786
[11]	AISI 4147	1.7225, SCM445	TM	869	1007	19	61	K	53,047	−0.2610	0.9374
[30]	AISI 4340	1.6511, SNCM439	TM	1234	1338	14	46	K	21,861	−0.1380	0.6326
[31]	SNCM439	1.6511, AISI4340	TM	1230 *	1344	15 *	45 *	K			
[32]	SNCM439	1.6511, AISI4340	TM	1147	1201	18	59	K			
[33]	X80	1.8758	F + B/AF	595	705	24	na	N _f	37,051	−0.400	0.9556
[34]	SCM435	1.7220, AISI 4130	F + P	360 *	540 *	17 *	78 *	N _f	122,510	−0.4230	0.9999

Table 1. Cont.

Ref.	Steel Designation		Micro-Structure *	Tensile Properties in Air				Mechanical Property Tested in H ₂	Fit Parameter		
	As Given in Ref.	Equiv. Grades *		YS, MPa	UTS, MPa	E _f , %	RA, %		m	n	R ²
[35]	AISI 1020	1.1151, S20C	F (+P)	207	379	na	na	da/dN	13,860	0.2203	0.0942
[36]	ASME SA105 Gr 2	1.0460	F + P	269	462	33	63	da/dN			
[37]	SM490B	1.0545, K12000	F + P	360	537	na	na	da/dN	15,878	0.1451	0.9248
[38]	SM490B	1.0545, K12000	F + P	360	540	17	78	da/dN	13,446	0.0851	0.5481
[39]	SCM435	1.7220, AISI 4130	TM	700	828	na	72	da/dN	6811	0.3035	0.9859
[23]	HY-100	1.6742, SFVQ 3	TM	758	855	25	70	da/dN	6130	0.1696	0.9456
[40–43]	Pure Fe	1.08xx, SUY	F	133	252	na	na	da/dN	26,060	0.0796	0.7922
[44]	X100	-	B	679	802	22	76	tensile RA	58,951	−0.2300	0.9394
[45]	X70/X80 heat B	1.8758	F + P	565	600	na	na	K	72,923	−0.1850	0.7091
[46]	X80 heat E	1.8758	F + P	593	710	36	na	K			
[46]	X80 heat F	1.8758	F + P	552	696	34	na	K			
[47]	X70	1.8758	F + P	584	669	20	57	K			
[48]	X80	1.8758	F + P	510	689	na	na	K			
[45]	X60HIC	1.8973	F + P	434	486	na	na	K	95,104	0.0570	0.9303

4. Discussion

As mentioned previously, the experimental data were fitted using Equation (2). For $|n| = 0.5$, all rate limiting steps (Figure 4) would be undistorted. However, the analysis performed in this study shows that the influence of hydrogen fugacity upon the mechanical properties of steels does not follow a $f^{[0.5]}$ law for a wide variety of microstructures, strength levels, and test methods (Table 1).

Figure 4. Rate-limiting steps for the transport of hydrogen to the fracture propagation zone. 1 = transport of hydrogen to the crack tip, 2 = physical adsorption, 3 = dissociative chemical adsorption, 4 = hydrogen absorption, 5 = transport of hydrogen to regions of tensile stress, 6 = hydrogen–material interactions.

The first part of the discussion focuses on the possible rate-limiting steps in the hydrogen reaction chain (Figure 4). It is known that for metals loaded in a hydrogen gas environment, hydrogen-assisted crack initiation typically starts at the gas–metal interface. Upon further loading, this crack propagates into the material until a final cross-section fails by overload fracture. In other words, special attention must be paid to the possible rate-limiting steps at the gas–metal interface.

As outlined in [8], the collision rate of gas molecules with a metal surface is linearly proportional to the gas pressure, and a sticking coefficient of approximately 1 can be assumed for pure metal surfaces. Under such conditions, around one monolayer of H_2 adsorbs within one second at a pressure of 10^{-10} MPa hydrogen partial pressure. With this estimation, it appears very unlikely that Steps 1 and 2 (the transport of hydrogen to the crack tip and physical adsorption) are rate-controlling under the test conditions discussed in this study with fugacities up to approximately 250 MPa.

The role of dissociative chemical adsorption was discussed in detail in [49]. It was shown that on Fe and Ni single-crystal surfaces, the coverage of dissociative sites is 100% at hydrogen gas pressures in the order of 10^{-1} MPa. Whether this is also true for highly distorted poly-crystals along a crack surface has not been investigated so far [8]. However, when considering fugacities up to approximately 250 MPa, it appears unlikely that dissociative chemical adsorption is rate-limiting under the test conditions discussed in this study.

The role of hydrogen absorption can be assessed by comparing test results obtained in gaseous hydrogen atmospheres with those of hydrogen pre-charged specimens. Hydrogen pre-charging of materials is typically performed under charging conditions to ensure a homogeneously saturated microstructure, and it appears that the fugacity-modified Sievert's law accurately predicts the hydrogen concentration in austenitic stainless steels pre-charged at high fugacities (Figure 5a). This method eliminates Steps 1 to 4 from the reaction chain (Figure 4). However, the analysis of data obtained with hydrogen pre-charged materials (Figure 5b) reveals $|n|$ values in the same range as obtained with tests in gaseous hydrogen atmosphere being significantly lower than 0.5. Since this trend does not change after eliminating Step 4 from the reaction chain, it is unlikely that hydrogen absorption is a primary rate-limiting step when testing in gaseous hydrogen atmospheres.

Figure 5. (a) Fugacity-modified Sievert's law with data from [50,51], (b) RRA of 304 L austenitic stainless steel [5] and relative fracture stress (RFS) of AISI 4135 martensitic high strength steel [7] pre-charged with hydrogen as a function of hydrogen concentration.

Although not fugacity-dependent rather than microstructure-dependent, the transport of hydrogen to regions of tensile stress, i.e., the crack tip, can also be rate-limiting. Hydrogen transport over distances larger than a few grain diameters is controlled by bulk Fickian diffusion, characterized by the diffusivity D . The diffusivities at room temperature are in the order of 10^{-15} to 10^{-16} m²/s for face cubic centered (fcc) austenitic stainless steels [9] and 10^{-8} m²/s to 10^{-11} m²/s for body cubic centered (bcc) steels depending on the microstructure [52,53]. Although the diffusivities of bcc steels and austenitic steels differ by several orders of magnitude, a significant difference in their mechanical response as a function of fugacity could not be identified (Figure 2). In other words, if D were a primary rate-controlling parameter (at temperatures around room temperature), the fugacity dependence of the HEI data of bcc and fcc austenitic steels should be clearly different, which is not the case. Therefore, it appears unlikely that hydrogen bulk diffusivity is a primary rate-limiting parameter when testing in a gaseous hydrogen atmosphere.

If Steps 1 to 5 of the hydrogen reaction chain are not rate-limiting, then the mechanisms involved in the hydrogen–material interactions (Step 6) must control the hydrogen embrittlement effects. The length scale is in the order of the plastic zone in front of a flaw or crack tip, typically around a few 10 to a few 100 μ m. Although the size of this zone comprises several grain diameters, the mechanisms are dominated by the so-called strain rate factor of the hydrogen transport equation, which is not a function of D [54–56]. There is experimental evidence that hydrogen influences dislocation motion and it can fairly be assumed that this effect increases with increasing local hydrogen concentration around the dislocation core. In this case, the interaction of hydrogen with dislocations plays a predominant role [8,57,58]. Unfortunately, currently, such mechanisms are not understood at the level of detail to quantify such effects. However, modern simulation tools may provide a framework to study such effects, since a model based on hydrogen accumulation and transport around a micro-crack using embedded atom methods and density functional theory calculations showed promising results [59].

The second part of the discussion focuses on the selection of material test fugacities (pressures) for high-pressure hydrogen applications. Figure 6a shows a histogram of the fit exponents ($|n|$) from Table 1, which suggests an asymmetric distribution. A statistical analysis revealed that these experimental data were well represented by a Weibull distribution. This appears reasonable because Weibull distributions are often found in statistics where values are bound to natural limits—here, $0 \leq |n| \leq 0.5$. The corresponding Weibull curves

are shown in Figure 6b with an expected $|n|$ value of approximately 0.12 and 95% of the $|n|$ values are lower than 0.37.

Figure 6. (a) Histogram of the fit exponents ($|n|$) from Table 1, (b) corresponding Weibull curves, scale parameter 1.727, shape parameter 0.195, coefficient of determination $R^2 = 0.895$.

The following discussion focuses on the two characteristic $|n|$ values, i.e., $|n| = 0.12$ and $|n| = 0.37$. To discuss the effect of these results on the definition of hydrogen test fugacities (pressures), two engineering applications were selected as use cases. The typical test pressure for automotive hydrogen tanks is 87.5 MPa ($f_1 = 155$ MPa) and the typical operating pressure of gas pipelines is 10 MPa ($f_2 = 11$ MPa). These nominal fugacities were used to calculate severity factors as

$$SF_{1,2} = \left(\frac{f}{f_{1,2}} \right)^n, f < f_{1,2} \quad (3)$$

using $|n| = 0.12$ and $|n| = 0.37$ (Figure 7). SF indicates how much less severe a test at f is compared to the reference fugacities f_1 and f_2 , respectively. For example, for $f_1 = 155$ MPa and $|n| = 0.12$, the severity factor of $SF = 0.94$ indicates that a fugacity of approximately 93 MPa is less severe by a factor of 0.94 compared to a fugacity of 155 MPa (Figure 7).

The severity factor of the fugacity can then be compared with the inherent variation in the measured mechanical property. Repeatability and reproducibility studies on high-pressure gaseous hydrogen's effects on the tensile properties of various structural materials [60,61] revealed that when only considering the scatter of quasi-static tensile test results in a control atmosphere, the RRA value must be lower than 0.89 to 0.99 (89% to 99%), depending on the material, in order to account for this reduction in hydrogen effects with a 95% confidence level (Table 2). The average value of 0.94 is represented by the solid horizontal line in Figure 7. Using again $f_1 = 155$ MPa and $|n| = 0.12$ as an example, the severity factor of $SF = 0.94$ indicates that a fugacity of approximately 93 MPa is less severe by a factor of 0.94 compared to a fugacity of 155 MPa. However, the measured hydrogen effects lie within the scatter of the results in the control atmosphere and, thus, are not statistically relevant with a 95% confidence level. In other words, for materials following an $f^{0.12}$ law, any hydrogen effects measured between 93 and 155 MPa are not statistically relevant with a 95% confidence level. The lower bound fugacities for $f_2 = 11$ MPa and $|n| = 0.37$ are given in Table 3. This analysis may provide a data-based rationale for defining appropriate test fugacities (pressures) for high-pressure hydrogen applications.

Figure 7. Severity factor (SF) quantifying the effect of hydrogen fugacity on hydrogen embrittlement indices assuming an f^n law with $f_1 = 155$ MPa and $f_2 = 11$ MPa as use cases. For example, for $f_1 = 155$ MPa and $|n| = 0.12$, the severity factor of $SF = 0.94$ indicates that a fugacity of approximately 93 MPa is less severe by a factor of 0.94 compared to a fugacity of 155 MPa.

Table 2. Results of repeatability and reproducibility studies of tensile tests in ambient control atmospheres (helium, nitrogen or air). RR = round robin.

	A286	Inconel 718	AISI 4340	SUS316L	1.4404
Tensile strength (MPa)	1166	1401	1374	710	587
No of labs participating in RR	6	6	6	3	1
No of specimens tested per lab	3	3	3	3	10
RRA lower 95% confidence	0.94	0.89	0.93	0.99	0.98
Reference	[60]	[60]	[60]	[61]	own results

Table 3. Calculated lower bound fugacities (pressures) for $SF = 0.94$.

$SF = 0.94$	$f_1 = 155$ MPa ($p_1 = 87.5$ MPa)	$f_2 = 11$ MPa ($p_2 = 10$ MPa)
$n = 0.12$	93 (62)	6.7 (6.4)
$n = 0.35$	131 (79)	9.5 (9.0)

5. Summary

The findings of this review can be summarized as follows:

- The relative reduction in mechanical properties in gaseous hydrogen experiments follows an $f^{|n|}$ power law (f = fugacity).
- The exponent $|n|$ was found to be smaller than 0.4. Since theoretical assumptions predict $n = 0.5$ for a fully undistorted hydrogen reaction chain, the reviewed data are interpreted in a way that hydrogen–dislocation interactions are the rate-limiting step controlling hydrogen effects in steels.
- A statistical analysis of the reviewed data revealed that 95% of the $|n|$ values are lower than 0.37.
- This analysis might be used to define appropriate test fugacities (pressures) to qualify materials for high-pressure gaseous hydrogen applications.

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