EPRG-PRCI-APGA
23rd Joint Technical Meeting
Edinburgh, Scotland
6–10 June 2022

LINE PIPE CAPABILITY ASSESSMENT FOR THE SAFE TRANSPORTATION OF HYDROGEN
PAPER NUMBER: 02

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ABSTRACT

The global energy shift towards lower carbon intensive energy sources is prompting the usage of hydrogen, transported via high pressure transmission pipelines. It is proposed that injection of a small quantity of hydrogen into the existing natural gas network will preclude the large-scale shift to hydrogen energy sources with high hydrogen blends or dedicated transmission pipeline infrastructure. This work is a compilation of published studies of the effect of hydrogen on pipeline steels, including vintage 1950s to modern steels. The embrittlement effect of hydrogen on static and fatigue properties is discussed, including consideration of the role of pipe strength, microstructure, and vintage. The pipe body, long seam weld, and girth weld data fatigue test data was collated and summarized and compared against the existing transmission pipeline network. Finally, recommendations for future testing programs are proposed to provide comprehensive understanding of hydrogen's effect on pipeline steels.

1. BACKGROUND

There have been hundreds of publications globally on hydrogen's impact on materials similar to common pipeline steels, from the initial interest in a hydrogen economy in the 1970s [1], [2] to the more recent global push for alternative fuels and renewed research [3]–[5]. This has culminated into the numerous roadmaps, studies or official statements from large organizations or governments across several countries [6]–[22]. These roadmaps or official initiatives often state certain 'safe' hydrogen blending concentrations however do not provide significant technical justification for these resulting partial pressures. It has been speculated that these safe blending concentration limits are more related to low-pressure downstream end-user limits (e.g., existing appliances, burners, etc.) and not necessarily justified for high-pressure transmission pipeline blending limits.

This paper summarizes the PRCI report (PR-214-214504), which condenses the global research on hydrogen's impact on line pipe specified by API 5L, CSA Z245.1, and other common materials used in transmission pipeline networks [23].

2. HYDROGEN'S IMPACT ON STEEL PIPELINES

It is well known for several decades that hydrogen can cause embrittlement in ferritic steels. Various embrittlement and damage mechanisms have been theorized and characterized by industry. This paper will not delve into these mechanics as Amaro et. al, PRCI, and others have proposed categories for hydrogen-assisted damage mechanisms: hydrogen enhanced decohesion (HEDE), hydrogen enhanced localized plasticity (HELP), and adsorption induced dislocation emission (AIDE) [24]–[28]. Regardless of the exact mechanism of hydrogen dissociation and resulting hydrogen concentration or flux through the steel, hydrogen will actively degrade a metallic material's mechanical properties.

When considering pipelines, the presence of hydrogen, even at small amounts, has shown a significant decrease in fracture resistance [29]–[31]. This would indicate that any hydrogen injection, even at small volumes and partial pressures, can have significant impacts on the performance of a pipeline system.

2.1. Static Loading

Static loading can be defined as a constant or slowly increasing force applied to the material (e.g., operating pressure, tensile test). In general, hydrogen has been observed to contribute little or no reduction in yield or tensile strength [32]–[34]. It has been stated by Boukortt et. al that the reduction in area (ductility) of various steels subjected to a hydrogen environment relative to air decreases with increasing yield stress [33]. The effect of strain rate on reduction in ductility has also been studied where slower strain rates showed a significant reduction compared to higher strain rates, shown in Figure 1 [34]. Chou reported that a strain rate reduction from 100-to-0.01 in/in/min resulted in a 4.5 times reduction in ductility in a presumably 1950's AISI 1020 steel, a material with similar chemistry and microstructure as line pipe steels of the era [35]. Duncan et. al also found that the weld metal and heat-affected zone in a A106 Gr. B material exhibited a 33-42% reduction in area [36].

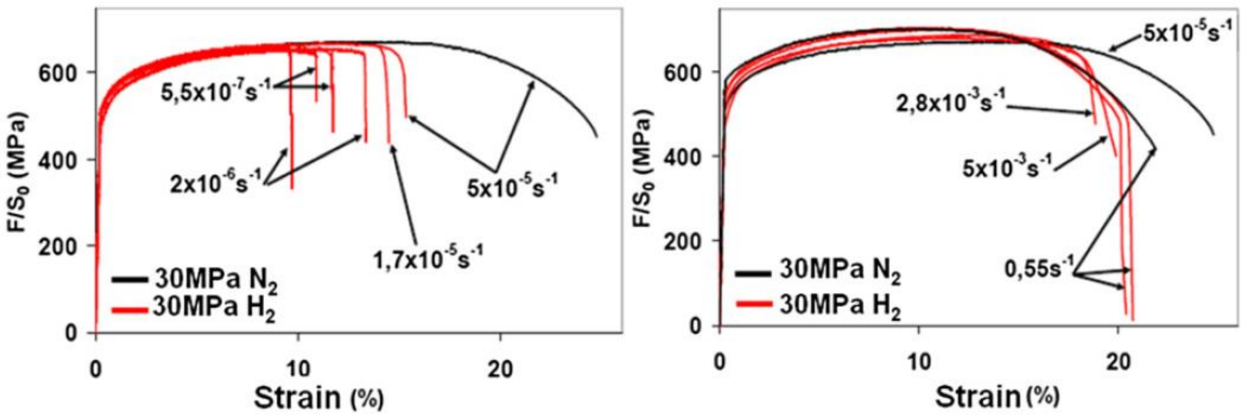

Figure 1 – Demonstration of Reduction in Ductility of an API 5L X80 Line Pipe Material Under Different Strain Rates [34].

In both Canada and the USA, nearly all pipelines are designed and constructed according to industry standards (e.g., CSA Z662, ASME B31.8) to operate below the line pipe steel's specified minimum yield strength (SMYS), with a maximum hoop stress of 80% SMYS from which additional safety margins are applied. The maximum combined stress from primary and secondary stresses is often limited to SMYS or lower. These factors indicate that hydrogen's negligible effect on yield strength is likely to not affect the conversion of existing pipelines or the design and construction of dedicated infrastructure when considering primary and secondary stresses based solely on SMYS. It should be noted that the reduction in ductility will likely reduce the inherent factors of safety built-in to these design codes. In specific scenarios, a limit states design that allows a pipeline to operate at stresses well within the plastic regime of a material may be employed to further design for and operate within challenging operating conditions. In both scenarios, hydrogen's reduction in plasticity should be considered.

Hydrogen's impact on the static tensile strength properties of pipeline steels was not further reviewed as the embrittlement damage mechanism will primarily dominate fracture toughness or fatigue loading scenarios, and therefore govern the long-term integrity considerations of a hydrogen-service pipeline.

2.2. Fatigue Loading

Recent studies in fatigue crack growth in hydrogen embrittled materials indicate that hydrogen diffusion rates ahead of an advancing crack tip can influence the rate of crack advance. Or alternatively, if the advancing crack growth rate (da/dN) is small, hydrogen diffusion ahead of the crack tip in the plastic zone will lead to higher crack growth rates compared to air due to hydrogen embrittlement [25].

In general, the effect of hydrogen on the fatigue crack growth rate of pipeline steels has been defined into three (3) regions, as depicted in Figure 2:

- Region A: at sufficiently low ΔK (e.g., < 8 MPa$\sqrt{\mathrm{m}}$), most ferritic pipeline materials do not experience measurable hydrogen embrittlement and the FCGR is similar to tests conducted in an inert environment.

- Region B: this "transient" region initially experiences a sharp increase in crack growth once a critical ΔK is achieved (e.g., approx. > 8 MPa$\sqrt{\mathrm{m}}$). The rapid increase in FCGR is thought to be a result of increased hydrogen diffusion to the crack tip plastic zone, resulting in higher crack growth rate per cycle and potentially correlating with the plastic zone size (i.e., zone of hydrogen embrittlement). A characteristic "knee" point is also observed within this region where at a particular ΔK, the FCGR is observed to decrease at increasing ΔK. This post-knee decrease in FCGR is thought to result from crack growth rates that are greater than the hydrogen diffusion and concentration into the plastic zone ahead of the crack tip. This would reduce the influence of hydrogen and degree of embrittlement, resulting in a decreased FCGR.

- Region C: at ΔK greater than the "knee" point, the behavior of the FCGR curve is similar to the FCGR curve of the material in air, albeit at a higher growth rate magnitude.

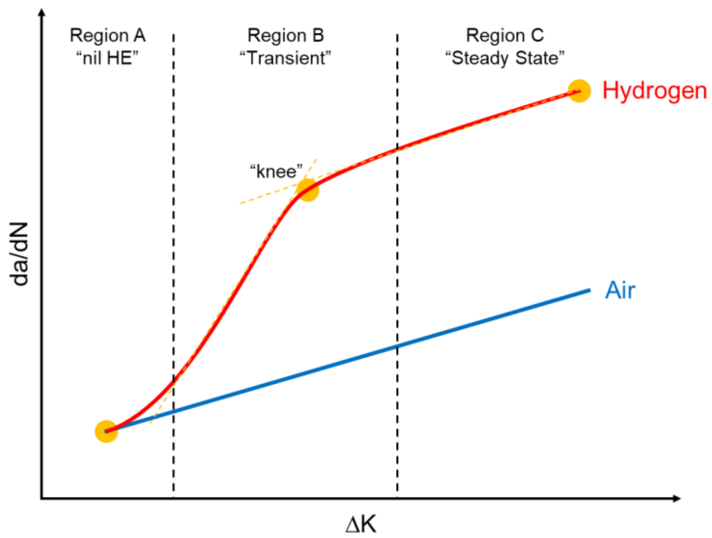

Figure 2 - Typical da/dN vs ΔK Curve for Air and Hydrogen Charged Steel Materials. Yellow dots indicate extracted data points.

Frequency and load ratio are key parameters when considering fatigue damage in hydrogen environments. The effect of loading frequency on a 1980s banded ferritic-pearlitic X42 steel was found to be negligible between 0.1-10 Hz [37]. It was also stated that hydrogen's effect on FCGR might be greater at frequencies less than 0.1 Hz, however, the resulting testing time is prohibitive but should be studied. Conversely, the trends are somewhat contradictory between 0.1 Hz and 10 Hz for a C-Mn steel [38] and X70 line pipe [39]. The effect of load ratio on fatigue has been determined where a load ratio of 0 (R=0) is found to be the most damaging (highest FCGR) [37]. This would imply that a pipeline that experiences routine shutdowns or depressurizations would experience higher fatigue crack growth than a pipeline that that experiences less significant pressure fluctuations or shutdowns.

Stalheim et. al found that a lower load ratio (0.1 vs 0.5) resulted in higher FCGR for modern X60 and X80 line pipe [40]. Conversely, Chen found that increasing load ratio from 0.1 to 0.5 increased the FCGR in two TMCP API X70 pipeline steels in a hydrogen environment [41]. Interestingly, a contradictory observation on the effect of load ratio was found between an X42 pipeline steel [37] and a different 1980s X42 steel [42].

The distinct role of frequency and load ratio should be considered when fatigue testing line pipe materials, balancing testing time and loading spectrums relative to the planned operations of the pipeline.

2.3. Role of Pressure and Temperature

It has been reported that hydrogen diffusivity of various girth weld electrodes and welding process showed distinct variations between each weld and a significant reduction in diffusivity compared to the base material [43]. As summarized by Hagen et al. from various researchers, the lattice concentration of hydrogen is often significantly higher than the traps and the traps were not observed to have any measurable contribution to the failure mechanisms of line pipe steels in hydrogen service [44]. Traps are useful however in absorbing diffusible hydrogen in welds, reducing the embrittlement effect, thereby reducing cold cracking upon cooling. They are only effective in transient hydrogen fluxes and do not provide benefit in long-term hydrogen service where the traps will be “full” and the matrix saturated.

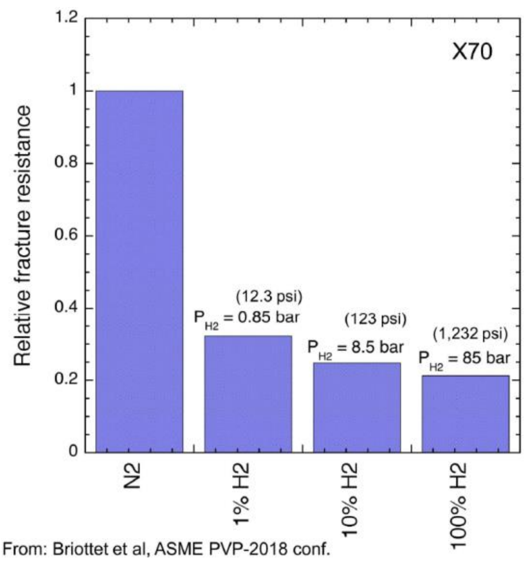

Figure 3 - Effect of Hydrogen Concentration on Fracture Resistance with Hydrogen Concentration and Pressure. Taken from [29] as adapted from [30].

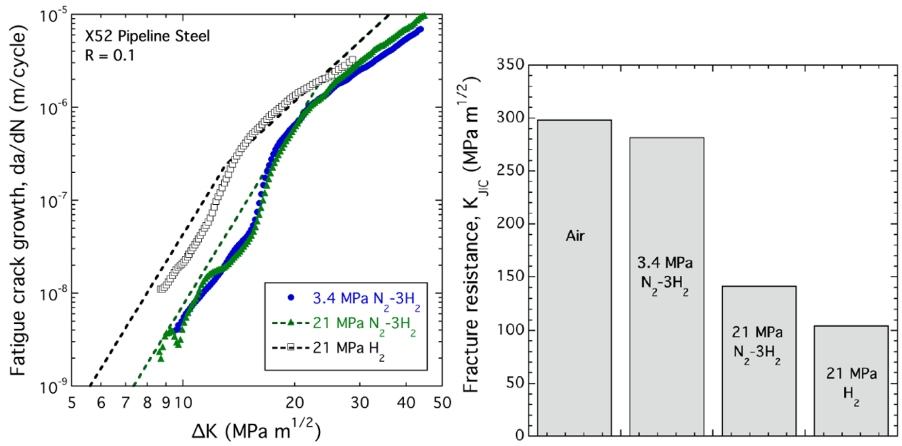

Figure 4 - Role of Hydrogen Concentration and Partial Pressure on Fracture Toughness and FCGR of a Modern X52 Material [31].

2.4. Role of Microstructures

Microstructure and strength can vary significantly between line pipe steels, based on their grade, manufacturing year, manufacturing method, and intended service conditions. The role of microstructure on the steel's resistance to hydrogen embrittlement has been the focus of numerous studies. The nature of steel microstructures and their extraordinary variability does present challenges when comparing studies to each other. This makes determination or ranking of the exact role of individual phases in hydrogen embrittlement difficult, however, some observations could be elucidated.

Some hydrogen performance studies from Sandia and other researchers (e.g., [45]–[48], [49]) were completed on a variety of materials, including SA 372 Grade J, Cr-Mo, Ni-Cr-Mo, and aluminum. These findings were excluded from this paper as these materials are not routinely employed in pipeline systems.

There have been many tested materials by Sandia National Laboratories as part of the Technical Reference for Hydrogen Compatibility of Materials study [50]. Across the pipeline steels studied, these could be categorized as 'transition' materials from the 1970-1990s where steels were generally cleaner, lower alloyed than vintage steels, and strengthened to a lesser degree of thermo-mechanical processing than modern steels. Furthermore, the Technical Database of material test summaries provided by Sandia (link) only contains data for a pre-1985 X42, a 2000s X60 with demonstrated hydrogen induced cracking (HIC) resistance, and several 2000s X80 line pipe steels. The report and database only contain a fraction of the total tested materials by Sandia or others and their usefulness is limited. A detailed summary of Sandia's numerous tests [29], [31], [37], [40], [45]–[90], National Institute of Standards and Technology tests [1], [24], [42], [91]–[106], and other researchers could provide a useful foundation for the pipeline industry.

Amaro et. al, when studying an API 5L X100 steel under 1.72-to-20.6 MPa of hydrogen pressure, observed the fatigue crack morphology and crack path. At lower ΔK and da/dN values, the crack growth was observed to be primarily intergranular. At higher ΔK and da/dN values, the crack growth was observed to be primarily transgranular. This indicates that different mechanisms are activated and dominate fatigue crack growth depending on crack extensions per cycle. For reference, the same material tested in air observed a mixed intergranular-transgranular character in the fine-grained and primarily bainite and acicular ferrite microstructure [24], [104]. Additionally, an API 5L X52 material comprised larger-grained ferrite and pearlite was tested under the same conditions as the X100 material as Amaro et. al [95]. When tested in air, the X100 and X52 steels exhibited the same fatigue crack growth mechanisms, however, when tested in a hydrogen environment, the X52 material had higher stress intensity (ΔK) and a higher initial fatigue crack growth rate than the X100 material. Additionally, Amaro et al. proposed a correlation between FCG (da/dN) and ΔK, further stating that it appeared that materials with higher proportions of polygonal ferrite experienced higher FCGR than materials with lower proportions of polygonal ferrite [94]. This observation occurred across several materials however the grades, phase balances, and material identification within the paper is not clear. The increased embrittlement of polygonal ferrite compared to acicular ferrite was also observed by Angus [107]. The degree of polygonal ferrite embrittlement is partially proportional to the grain size, where larger grains increased the tendency for transgranular fracture; however, at very small grain sizes, the increased grain boundary area is thought to increase the preferential segregation of hydrogen at the boundaries and increased embrittlement.

Chen studied two different API 5L X70 pipeline steels with modern low-C chemistries and microstructures under various load ratios and fatigue frequencies [41]. The material with a small amount of bainite (~2%) in a primarily PF/AF microstructure exhibited a higher FCGR than the other material with a PF/AF microstructure. It should be noted that Chen's experimental details (such as

YS/TS of the two materials, baseline material toughness, etc.) and hydrogen environment information is sparse, making this observation somewhat unreliable and potentially contradictory to Amaro's observation. It has also been reported that bainitic structures likely experience a higher degree of embrittlement as the hydrogen accumulates along the bainite laths in API 5L X80 and X100 steels [108]. In an X120 material, HIC susceptibility increased with the amount, area, and volume fraction of inclusions, in addition to increased susceptibility of granular bainite and M/A phases [109], [110]. Briottet et. al observed that an API 5L X80 material fractured in a brittle manner with quasi-cleavage fracture appearance, with delamination and micro-cracking occurring along pearlite alignments [34]. Park et. al studied the influence of different microstructures in the same X70 material and determined that a F/AF structure was more HIC resistant than a F/B structure, likely due to the higher toughness and prevention of accumulation of M/A phases [111].

Chatzidorous et. al studied the base metal and long seam HAZ performance of a 1970s X52, an unspecified manufacturing year X65 (BM only), and a 2000s X70 pipeline steel in an aqueous charging environment [112], [113]. The X52 material was comprised of a banded ferrite-pearlite structure. SEM analysis of the fractured surface displayed distinct morphologies between the ferrite and pearlite bands and no microcracking was observed. The X70 base material comprised of a banded ferrite and mixed pearlite/bainite structure. Several MnS and Al_2O_3 inclusions were located at the banding interfaces. The HAZ was comprised of a similar microstructure but with more equiaxed grains. The presence of boundary inclusions and the hardness (and by association, strength) differences between the banded phases was the likely cause for the observed microcracking in the X70 material. In contrast to the banded structures of the vintage X52 and modern X70, the X65 steel possessed a uniform microstructure comprised of a primarily ferritic-bainitic structure with intermittent pearlite islands and the presence of martensite-austenite (M/A) islands. The M/A islands and higher dislocation density of the bainite phases is attributed to the increased hydrogen damage compared to the banded X52/X70 materials. This would imply that the banded ferritic-pearlitic microstructures are more resistant to hydrogen than uniform ferritic-bainitic microstructures.

The effect of microstructural texture (banding) on the FCGR was studied on a 90% polygonal ferrite and 10% ferrite API X65 steel by Ronevich et. al, summarized in Figure 5 [61]. The 508 mm diameter and 25.4 mm thick pipeline had an elongated banding in the pipe longitudinal direction. The notch orientation relative to rolling direction had a noticeable effect on FCGR. At $\Delta K < 15$ MPa$\sqrt{\mathrm{m}}$, the FCGR was significantly lower with the notch transverse to the rolling direction (L-R direction) compared to notches parallel to the rolling direction (C-L and L-C direction). When $\Delta K > 15$ MPa$\sqrt{\mathrm{m}}$, the crack growth rates of the parallel notches were still twice that of the transverse notch direction. Transverse notches were observed to possess significant branching and a tortuous crack path.

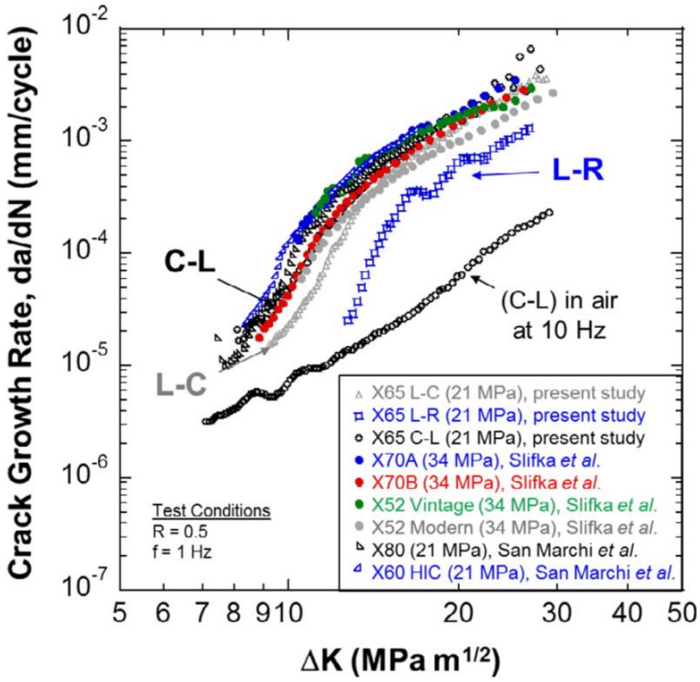

Figure 5 – Compilation of FCGR Across Multiple Steels. Banded microstructures: X52 vintage (70% PF, 30% P), X65 (90% PF, 10% F). Uniform microstructures: X70A (90% PF, 10% AF), X70B (90% PF, 10% B), X52 modern (90% PF, 10% AF), X80 (90% PF, 10% coarse AF), X60 HIC (100% PF) [61].

Wachob and Nelson studied the fatigue resistance of a A516 C-Mn pressure vessel steel subjected to three different heat treatment conditions (normalized, bainitic, martensitic) and two different prior-austenite grain sizes (200 µm and 30 µm) [114]. The fracture initiation threshold was lowest for the normalized (ferrite-pearlite) steel, increased by ~10% for bainitic tests, and ~20% for the martensitic tests. The fracture initiation threshold increased with increasing grain size. For all microstructure and grain size tests, at higher stress intensities and FCGR rates, all tests converged.

The role of individual microstructural phases appears to not definitively affect FCGRs. In general, polygonal ferrite has been observed to have a poorer performance than other phases, and normalized heat treatment conditions resulted in greater fatigue crack growth rates. It should be noted that identifying and ranking distinct phases based on their performance in hydrogen is difficult. Variations in mechanical processing, chemical composition, microstructural texture, testing direction, residual stress, and impurities/imperfections can influence the crack growth behavior. A relative ranking was proposed by Barthelemy and Pressouyre in 1985, shown in Table *1*, that appears to align with other research and can be useful in assessing vintage materials.

Table 1 - Summary of Rankings of Importance (damage) in C-Mn Steels in a Hydrogen Environment [115].

WHAT TO ASK FOR :	IMPORTANCE
- Quench and Temper microstructures, rather than normalized ones.	***
- Good "cleanliness" (S < .002wt%; O_2 < 40ppm)	***
- A steelmaking process that limits segregated zones; otherwise, ask for low "segregating" contents (P,Sb,Mn,...)	***
- IF possible: an addition of finely dispersed strong hydrogen traps (fine carbides,e.g. VC; atoms: e.g. Hf)	**
- Fine prior austenite or ferrite grain size (through appropriate thermomechanical treatments or carbide forming additions).	*
- Surface homogeneous compressive stresses (e.g. shotpeening)	*

*** : very important ** : important * : less important.

WHAT TO AVOID	IMPORTANCE
- As quenched martensite or bainite (e.g. such as in heat affected zones, or in segregated zones)	***
- Inclusion stringers (MnS, oxides)	***
- Large carbides (TiC,ZrC)	**
-"Segregating" elements (metalloïds e.g. :P,Sb, As; Mn,Si)	**
- High nickel content (C-Mn steel)	*
- Applied or residual local stress concentrations and/or deformations	*

2.5. Seam Weld and Girth Weld Performance

Current and future gaseous product transmission and distribution pipelines will always possess welds, whether in the long seam from line pipe manufacturing or circumferential girth welds from construction. There are a limited number of hydrogen service investigations that focus on welds, including weld metal and heat-affected zone.

According to Che's summary of several reports that are unavailable to the authors [116], the following items were stated. It should be noted that the three studies summarized below were completed in the 1980s and would likely be representative of the materials and practices of the 1970-1980s.

- The weld metal and HAZ of a girth welded X60 line pipe was observed to have the same FCGR as the base metal between ΔK of 10-20 MPa$\sqrt{\text{m}}$ at 6.9 MPa H_2 pressure, 1 Hz frequency and load ratio of 0.15 [117].

- FCGR of weld metal and HAZ was found to be affected by the magnitude of ΔK in a 550 MPa tensile strength steel in a 3 MPa H_2 environment and 0.1 Hz frequency [118]. When ΔK < 13 MPa$\sqrt{\text{m}}$, the FCGR of weld metal, HAZ, and base metal are similar. At higher ΔK, the weld metal and HAZ exhibited a higher FCGR than the base metal.

- In pressure vessel steels, a similar FCGR was observed at higher growth rates (>10^{-5} mm/cycle) and higher load ratios [119]. Hydrogen increased the FCGR near the threshold (ΔK_{th}), and reduced the threshold values, especially for the HAZ. Pressure vessel steels generally share similar metallurgy and mechanical processing as vintage line pipe.

Drexler et. al conducted an extensive study of a 1964 X52, 2011 X52, and two 2005 X70 pipeline steels, including their long seams and girth welds [93], [96], [103], [104]. Within the pipeline industry, the vintage term is associated with pre-1970s line pipe, differing from the authors use of the "vintage" term for any line pipe manufactured before 1990. Chen presumably provided additional details on the same alloys [97]:

- 1964 API 5L X52, 914 mm OD x 10.6 mm WT: cutout from Pacific Gas & Electric (PG&E).
- 2011 API 5L X52, 508 mm OD x 10.6 mm WT: Air Liquide special order material intended for hydrogen gas service.
- 2005 API 5L X70, 914 mm OD x 18 mm WT: line pipe steel intended for natural gas service, from CRC Evans.
- 2005 API 5L X70, 914 mm OD x 22 mm WT: experimental alloy intended for enhanced ease of welding, from El Paso Natural Gas company.

Observations from the works of Drexler et. al [103] and Chen [97] for these alloys revealed that the:

- Vintage 1964 X52 girth weld shows signs of an inside diameter (ID) root pass deposited after the completion of the weld joint, which is not typical/common for the era or technique (SMAW). Only modern mechanized GMAW sometimes have an ID root pass. Therefore, depending on the initial crack location in the test, the initial FCGR may not be representative of typical GW of the era as this weld is not representative of field welds. From the stated specimen geometry, it is likely that this ID weld pass, and it's associated reheat zones, was included in the fatigue crack.
- Vintage X52 long seam and girth weld HAZ exhibited higher FCGR, whereas the girth weld metal had a lower FCGR than the BM at lower ΔK. The modern X52 girth weld metal and HAZ exhibited lower or similar FCGR than the BM.
- X70 long seam weld and HAZ exhibited a higher FCGR than the BM, whereas the girth weld metal and HAZ exhibited a lower FCGR.
- Drexler et al. stated that the response of HAZ and weld metal was not consistent between the materials studied. This included a horizontal da/dN vs ΔK behavior at low ΔK in the HAZ for some alloys.

While the tests completed did follow a rigorous test matrix, Chen reported that tests were omitted from certain alloys due to a number of observations for the sake of reducing testing time. For example, it was reported that the X70 welds (not specified if GW or long-seam) were significantly more resistant to FCG compared to the base material and were omitted from further testing. However in general, the girth welds exhibited a similar or slightly higher FCGR compared to the base metals.

Slifka et. al studied the FCGR of girth weld and long seam HAZs of a 1964 X52, a modern X52, and two modern X70 materials [100]. These are the same materials studied by Drexler and Chen, however, with additional studies completed in the HAZ. At ΔK less than the "knee" point, the long seam HAZ FCGR was 1-2X higher than the base material. The girth weld metal exhibited a similar FCGR compared to the base material. When the hydrogen pressure was increased from 5.5 MPa to 34 MPa, the HAZ and girth weld metal exhibited a 2-5X increase in FCGR.

Ronevich et. al studied several welds:

- Five high-strength weldments for a 1990s X100 steel, with one (1) being an experimental friction-stir weld that is not applicable to cross-country pipelines [62], [64], [67], [120]. The

1990s girth weld, using unspecified welding consumables or welding parameters, produced weld metal FCGRs higher than the BM whereas the HAZ produced lower FCGRs. Using the same 1990s X100 base metal, additional GMAW welds were completed around 2020 with ER100S-G (matching), ER120S-G (overmatching), and a proprietary low-temperature phase transformation (LTPT) electrode intended for military armor applications [64].

  - ER100S-G and ER120S-G weld metal FCGR was similar or higher than the base metal.
  - HAZ performance was provided only for the unspecified 1990s girth weld. The HAZ FCGR was lower than both the BM and weld metal by a factor of 5.
  - The LTPT electrode, intended to reduce weld hydrogen cracking, possessed significant martensite formation, a displacive phase transformation that results in generating compressive stress within the weld metal. The LTPT electrode FCGR curve was very irregular, a function of the varying and significant compressive stress throughout the weld thickness from the intentional displacive phases.
  - These high-strength welds possessed a greater FCGR than the base material, and greater than lower strength girth welds [55].

- In a study of GMAW-welded X65 and a friction stir welded X52, the FCGR of the primarily acicular ferrite X65 weld and the polygonal ferrite X52 friction stir weld were very similar [55], [56]. Closer analysis of the X65 fusion zone and HAZ results indicated crack-closure occurring at lower ΔK values. When correcting for crack closure, the HAZ exhibited a higher FCGR [57], [61].
- Ronevich in [121] provided a summary of tested materials (modern & vintage X52, X65, X100), in coordination with Slifka et. al's work [100]. The fracture toughness (K) was stated for most of the base metal and weld locations. Interestingly, the vintage X52 exhibited several so-called "large inclusions" (~10-20 µm in diameter) in the weld metal by the authors, finding that they significantly decreased fracture toughness.
  - The authors are noting that although Slifka has labelled the vintage X52 weld as being completed by GMAW, it is highly unlikely that a 1964 pipeline girth weld was completed by GMAW but is likely a SMAW weld based on the construction practices of that era.

Benoit et al. studied the GW weld metal performance of a 1972 X60 and a modern L485 (X70) material [122]. The 1972 X60 was welded with cellulosic SMAW electrodes, while the modern L485 material was welded with GMAW that utilized different root and fill pass electrode chemical compositions. Weld metal microstructures and chemical compositions were provided however no weld hardness or tensile data is stated. In an 8.5 MPa gaseous environment with 25% H_2 (partial pressure of 2.1 MPa), testing revealed a 30% reduction in CTOD for the 1972 X60 weld metal whereas the modern L485 weld exhibited a 50% reduction. The X60 girth weld was further studied for its fatigue performance and significant variability in FCGR was observed. The author attributed the variability to localized regions of compressive residual stress in the weld metal, and observed a sharp acceleration in FCGR corresponded with an ~200 µm smooth volumetric imperfection. This imperfection would be considered acceptable by pipeline construction non-destructive examination (NDE) workmanship standards. Other research from Chen [41], Olden et. al [123], Davani et. al [124], Lee et. al [125], Holbrook et. al [37], and Mucci [126] provide additional studies on various alloys.

3. SUMMARY OF PUBLISHED FATIGUE TESTING

It was stated in 1979 by NIST/Sandia [1] researchers that:

> *… many investigators have found that mild steels do not display the dramatic changes in properties often associated with hydrogen effects and thus a rank ordering of materials and microstructures is not a straightforward task*.

This statement has since been further complicated by the development and significant usage of thermo-mechanical processing of lean alloyed, high strength steels in the past few decades by the pipeline industry.

The aforementioned sections highlight the volume of testing information that is available just for line pipe steels and distilling the findings into useful rankings in the absence of a rigorous design of experiment can make comparisons between studies difficult. To summarize the collected data, a dataset was created to collate basic pipe information and test results. While there are FCGR summaries available (e.g., [41], [127]), there is no collation of key pipeline attributes that might enable direct comparison of lab results to an operating pipeline or pipeline network.

The pipe information collected was aimed at gathering key information to allow operators to align their pipeline network to the test data to support a change of service engineering assessment. These variables include:

- Specification (e.g., API 5L, CSA Z245.1) and grade.
- Manufacturing year, mill, and seam type.
- Diameter, thickness, uniform elongation, and yield/ultimate tensile strength (transverse and longitudinal direction).
- Pipe body toughness (CVN, CTOD, K, J).
- Chemical composition (incl. API 5L IIW/Pcm, CSA Z245.1 formulas) and microstructure.
- Long seam weld properties (WM and HAZ YS/UTS, uniform elongation, toughness).
- Girth weld properties (process, electrode, WM and HAZ YS/UTS, uniform elongation, toughness).
- Test pressure (total pressure, hydrogen partial pressure) and testing standard.

The dataset had the goal of providing a thorough summary of key line pipe steel data to allow comparison of test results to the pipeline network. The published literature contained significant gaps in reported line pipe information (e.g., mill, manufacturing year), providing further challenges in understanding a particular materials performance or behavior. As discussed in Section 2.1, the primary focus is the fatigue performance of various pipeline alloys. As a result, the dataset focusses on the fatigue tested materials and does not fully document the negligible influence of hydrogen on yield and ultimate tensile strength properties. The dataset is summarized in Table 4.

3.1. Base Metal/Pipe Body

The base metal (pipe body) FCGR across all materials studied is shown in Figure 6. The trends are similar to those provided by Che [116] however more materials are included here. Additionally, this report demonstrates more variability in the base metal (pipe body) FCGR results with significant overlap at low stress intensities.

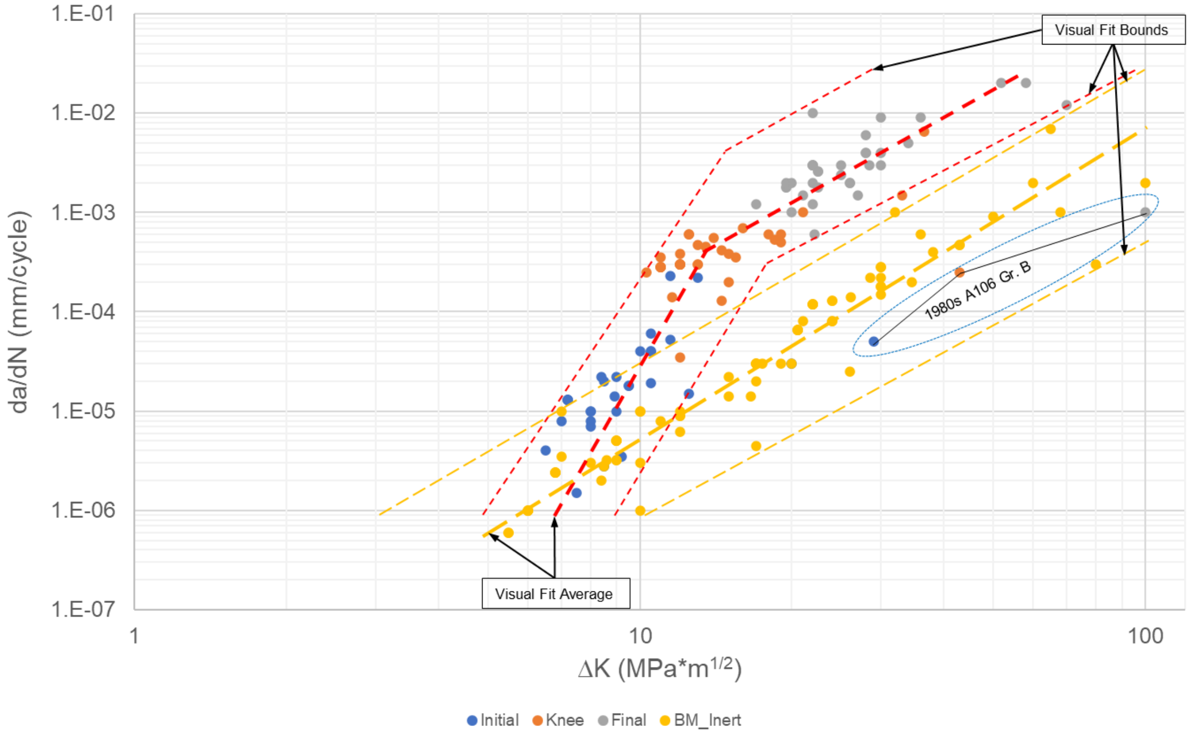

Figure 6 - Summary of Base Metal (Pipe Body) FCGR Results

There are a variety of published FCGR relationships for carbon steels (BS 7910 2019 edition) and line pipe specific steels (PRCI, [128]). These published relationships also include a two (2) standard deviation variation based on the tested materials, summarized in Figure 7. The BS 7910 and PRCI mean relationships agree well with the inert environment data collected in this study. Interestingly, the inert FCGR data collected in this work exhibited significantly greater variability than the BS 7910 and PRCI data. While the reason for the greater variability is not clear, it is possible that the variations were the result of testing completed over several decades and many researchers/testing facilities. In contrast, the PRCI testing data and relationships were created from a single testing facility.

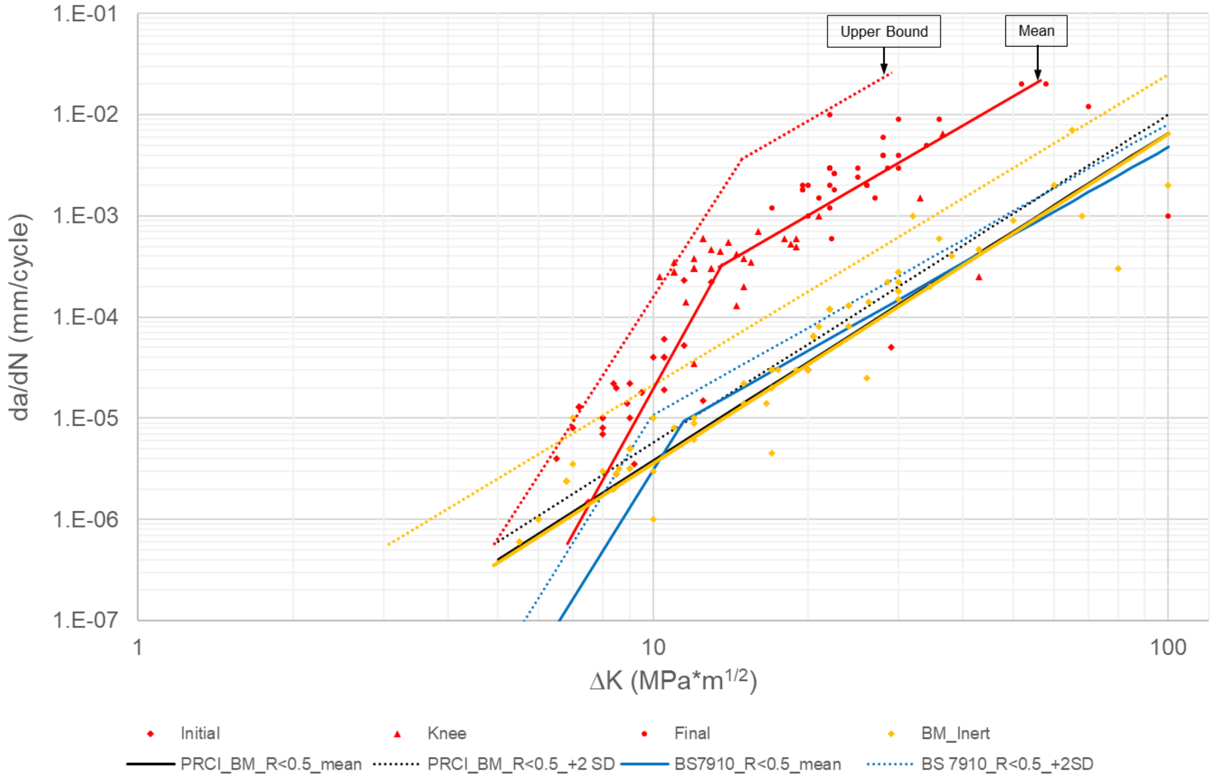

Figure 7 - Comparison of Base Metal (Pipe Body) Inert and Hydrogen Environment FCGR Data Including BS 7910 and PRCI [128] relationships.

Along with the FCGR curves, several reports have proposed Paris fatigue constants [3], [34], [39], [94]. Others have proposed the creation of S-N fatigue damage curves [129].

3.2. Long Seam and Girth Weld

The girth weld and long seam weld metal and heat-affected zone FCGR test results are summarized in Figure 8. There is significantly higher variability in the test results, further compounded by the lower number of data points. This variability is further exemplified when comparing the inert and hydrogen environment data points found in this report to those obtained from the PRCI study (Figure 9).

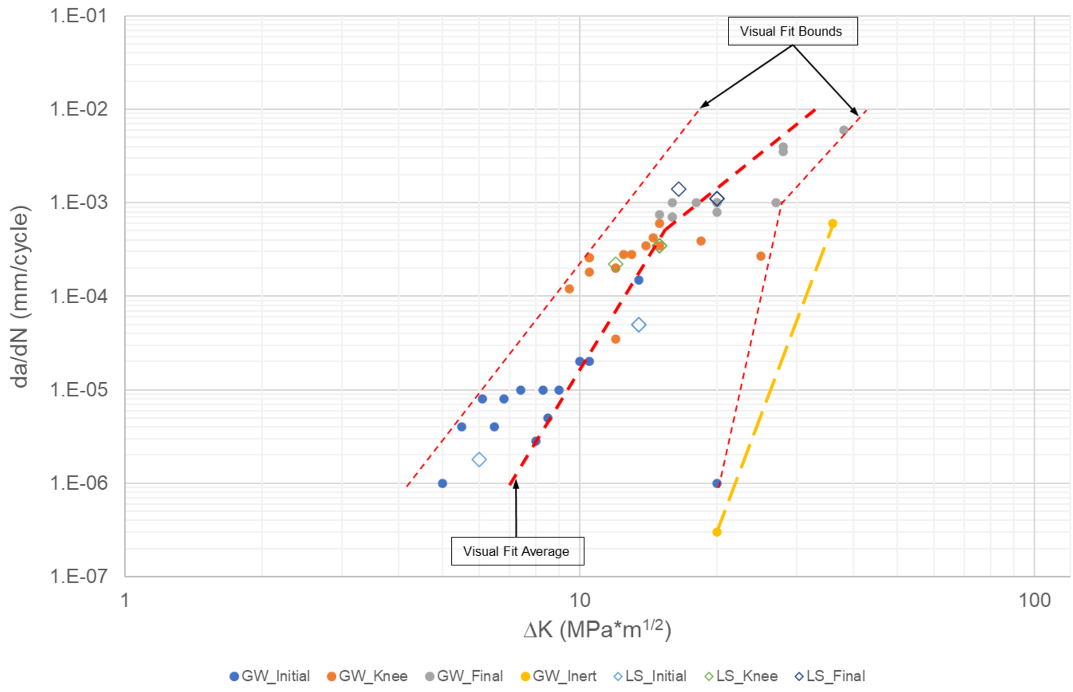

Figure 8 - Summary of Girth Weld and Long Seam HAZ and WM FCGR Tests

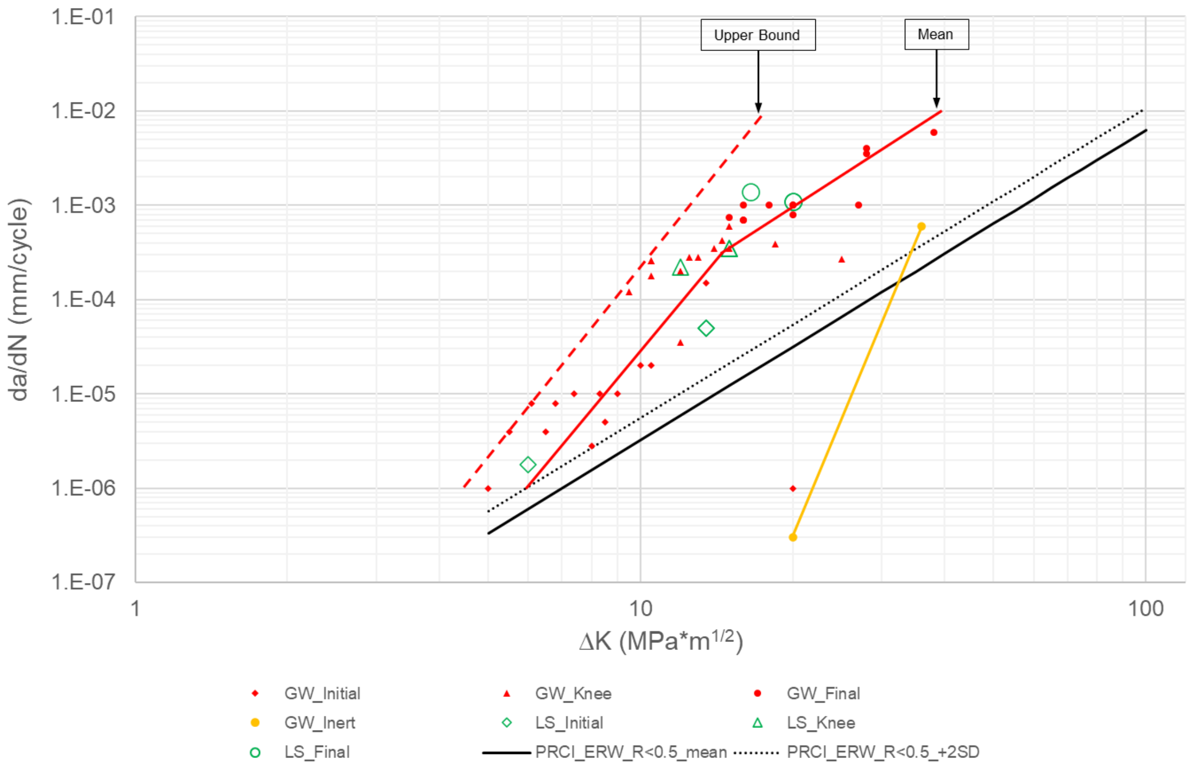

Figure 9 - Comparison of Girth Weld (GW) and Seam Weld (LS) Inert and Hydrogen Environment FCGR Data Including BS 7910 and PRCI [148] relationships.

4. NORTH AMERICAN PIPELINE NETWORKS

The US onshore gas transmission pipelines network has been constructed over several decades, with significant mileage constructed prior to 1970, shown in Figure 10. The majority of installed pipe is pre-1970s (vintage) and X52 or lower grades. Accordingly, the majority of installed line pipe steel is likely a banded ferrite/pearlite structure produced from rolling rather than the more homogeneous structure of modern materials.

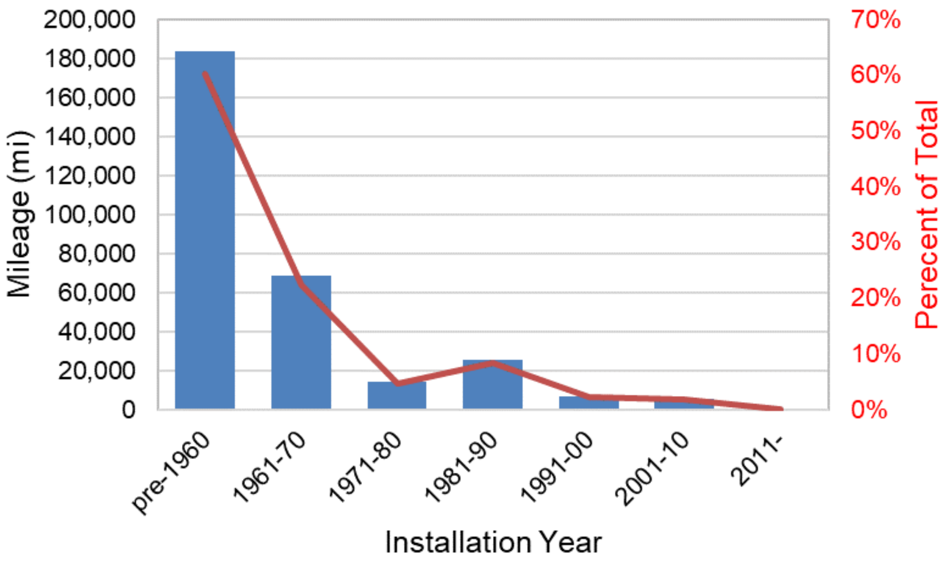

Figure 10: Summary of Gas Transmission Pipeline Mileage [source: US Bureau of Transport].

Pipe manufacturing methods have evolved significantly since the 1950s, capable of producing pipes of varying microstructures and mechanical properties. A high-level summary of line pipe manufacturing is listed below:

- 1950-1970s:
    - Primary strengthening mechanism was solid solution strengthening via alloying elements (primarily carbon, manganese).
    - Mills were capable of hot-rolling steel sheets, and typically utilized cold rolling to obtain desired plate thickness.
    - Microstructures are typically banded, with alternating parallel ferrite and pearlite phases.
- 1970-1990s:
    - Primary strengthening mechanism is a combination of solid solution strengthening (lower carbon and manganese, beginning of purposeful additions of chromium, molybdenum, and others) and thermomechanical work.
    - Mills used a combination of hot-rolling and inter-critical rolling with controlled cooling and strain rates to promote dynamic recrystallization.
    - Microstructures might be banded or homogeneous.
- 1990s and newer:
    - Primary strengthening mechanism is thermomechanical and cold-work, with small additions of titanium, niobium, and boron providing nucleation sites for initiation and refinement of phases (e.g., acicular ferrite).
    - Mills used a combination of hot-rolling and inter-critical rolling with controlled cooling and strain rates to promote dynamic recrystallization.
    - Microstructures are homogeneous.
- The manufacturing of seamless pipe has not experienced the degree of change in manufacturing methods as welded line pipe. Seamless pipes rely on a combination of solid-solution strengthening and thermomechanical work, generally producing a material with minor banding or homogeneous microstructures.

In addition to the complexities of pipe manufacturing, each pipe mill will have unique rolling and thermal capabilities (heating & cooling), producing variations in microstructure and strength between each pipe mill. An excellent summary of historical pipe mills is available [130].

The effect of near-yield plastic behavior, specifically a strain-hardening (roundhouse) or a yield-plateau (Lüders band) material, should also be considered. Many researchers have not stated if the tested materials exhibited a yield-plateau or roundhouse-style hardening behaviour.

In contrast to the natural gas transmission system, Bedel provided an overview of existing hydrogen pipelines in world [131]. While many pipelines show decades of presumably incident free operation, the operating pressures and corresponding hoop stress appear to be lower than a typical transmission pipeline although insufficient pipeline thickness data was provided. For example, the AGEC pipeline operates at 37% SMYS hoop stress or lower, nearly half of modern pipeline pressures thereby reducing hydrogen damage.

Cross country transmission pipelines are often constructed with welded line pipe, utilizing either EW (ERW or induction welded) or SAW (longitudinal or spiral/helical) long seams, and circumferential girth welds joining individual ~8-15 meter (~20-45 feet) pipe segments to each other. Long seams are welded in the pipe mill production facility under very controlled conditions. Girth welds are very often completed on the right-of-way under a variety of conditions, frequently resulting in larger variations in mechanical properties and imperfection types and frequencies than long seams.

Pipe long seams often have one or two weld passes. In EW pipe, a singular "weld pass" is formed when the abutting edges of the skelp plate are pressed together and fused, resulting in a small thin through-thickness weld perpendicular to the plate surface. In SAW pipe, it is common to utilize a two-pass weld design where the first pass is completed from one side of the pipe surface (e.g., ID) and the final weld pass completed from the other surface (e.g., OD). It is common practice in modern EW pipes to conduct a post-welding heat treatment or normalization of the weld region to improve mechanical properties. Girth welds are almost always completed using multiple successive weld passes (i.e., multi-pass weld), starting at the inner diameter ("root" pass) and additional weld passes are added to complete a full thickness weld that is "capped" with a final weld pass that often protrudes slightly above the pipe surface. Historically, the majority of North American transmission pipelines have been constructed by manual SMAW welding using cellulosic welding electrodes. Only recently installed (~2000 and newer) relatively thick and large diameter pipelines are likely to be welded with GMAW processes and a narrow gap joint design.

Common pipeline construction and welding codes require a minimum girth weld NDE inspection frequency and define the weld defect acceptance criteria for those inspected welds. These requirements are summarized in Table 2. As shown, unless a pipeline has been constructed with supplementary frequency and defect acceptance criteria, a significant portion of girth welds remain uninspected and could possess defects greater than the acceptance criteria. Furthermore, the stated defect sizes (width and length) are significantly larger than the "large" imperfections (e.g., 0.2 mm and 0.02 mm, [121], [122]) found to influence the FCGR test results. Additionally, the most common form of girth weld inspection is radiographic testing (RT) which cannot measure the through-thickness height of the imperfection.

Table 2 - Summary of NDE Inspection Frequency and Girth Weld Acceptance Criteria for North American Pipelines.

<table>
<tr><td colspan="2">NDE Inspection Frequency[1]</td></tr>
<tr><td>ASME B31.8</td><td>Minimum of 10%-75% of welds based on Location Class</td></tr>
<tr><td>CSA Z662</td><td>Minimum of 15% of welds</td></tr>
<tr><td colspan="2">Workmanship NDE Acceptance Criteria[1]</td></tr>
<tr><td>ASME B31.8 & API 1104</td><td><u>Radiographic Testing (RT)</u>
- Planar: 25-50 mm length
- Volumetric: 1.6 x 50 mm (elongated slag), 3 x 13 mm (isolated slag), 3 mm or 25% WT (isolated porosity)
<u>Ultrasonic Testing (UT)</u>
- Planar: 25% WT x 25 mm (surface-breaking) or 50 mm (buried)
- Volumetric: 3 mm (isolated) or 13 mm (cluster)</td></tr>
<tr><td>CSA Z662</td><td><u>Radiographic Testing (RT)</u>
- Planar: 12-50 mm length
- Volumetric: 1.5 x 50 mm (elongated slag), 1.5 x 12 mm (hollow bead), 2.5 or 33%WT x 10 mm (isolated slag), 3 mm or 25% WT diameter @ 3-5% area (porosity)
<u>Ultrasonic Testing (UT)</u>
- Planar: 3 mm or 25% WT x 25 mm length (surface-breaking) or 50 mm (buried)
- Volumetric: 3 mm or 25% WT x 50 mm</td></tr>
<tr><td colspan="2">[1]: Requirements taken from most recent code edition for example only. Code edition in effect at time of construction might have different requirements.</td></tr>
</table>

A recent phenomenon of significant heat affected zone softening has been observed in modern high strength (≥ X60) pipeline girth welds, sometimes resulting in pipeline failures [132], [133]. The softening has resulted in a decrease in yield and ultimate tensile strength. The effect of hydrogen on HAZ softening should be further investigated as HAZ softening has been demonstrated to occur at regions further away (inter-critical HAZ) from typical HAZ study locations (CG-HAZ). Hydrogen has been shown to concentrate at strained lattice locations, potentially leading to greater hydrogen damage at the softened zones.

5. CRITICAL COMPARISON OF TESTING TO-DATE

While there has been significant research in the past several decades on hydrogen's effects on line pipe steels, the number of studies and data points for relevant materials must be statistically significant and well quantified. Given the recent gas pipeline industry requirements for generating traceable, verifiable, and complete data records for operating pipelines (e.g., PHMSA CFR 192 Gas "Mega Rule"), it would be likely that such a similar and more rigorous requirement be placed on pipelines being considered for hydrogen service. This might include determination of a pipeline's distribution of strength, toughness, quantification of existing imperfections, or determination of strains greater than yield strength. As such, the fatigue testing that will form the technical foundation for operation of hydrogen pipelines needs to provide a statistically significant basis of representative pipeline materials for both operating pipelines and new yet-to-be constructed infrastructure.

When considering the studies reviewed in this work, only 22 unique materials were studied over the past four decades on hydrogen's unique effects on fatigue, summarized in Table 3. Of the limited published manufacturing years, only 1 material could be confirmed as vintage (pre-1970s) and many did not state the manufacturing year of the material studies. Most of the fatigue tests have been completed post-1990s and it could be inferred that a majority of the fatigue tests are on post-1990s materials.

Table 3 – Count of Fatigue Studies Completed on the Pipe Body and Welds

Grade	No. of Unique Materials	Manufacturing Years
Gr. B	1	N/A
X42	2	N/A
X46	0	N/A
X52	4	1964, 1990s, 2000s, 2011
X60	3	1972, 1980, N/A
X65	1	N/A
X70	3	2005, N/A
X80	3	N/A
X100	2	1990s, N/A
Other	3	N/A

Given the increased FCGR of welds, and specifically the heat-affected zone, a thorough understanding of welded long seams and girth welds is required. Two (2) line pipe long seams were studied, one from a 2011 X52 ERW and the other from a 2005 X70 DSAW pipe. Seven (7) girth welds were fatigue tested, with two (2) completed with SMAW (with 1 specified as cellulosic) and the remainder as completed with GMAW. Similar to the grade breakdown in Table 3, the number of tested long seam and girth welds is insufficient.

As discussed in Section 4, the existing pipeline network is comprised of widely varying pipe mills, manufacturing years, long seam types, girth welds, and overall quality (e.g., vintage LF-ERW). Given the large variability and resulting number of permutations for testing each combination at a statistically significant level, which would be very costly and time consuming, a strategic testing program should be considered. The program should balance the number of tests with pipelines that are being considered for hydrogen service to reduce unnecessary testing. For example, if an early 1950's pipeline, that is currently operating under significant pressure reductions from existing damages, might not be economically feasible to increase the operating pressure to offset the lower volumetric heat content when injected with hydrogen and therefore should not be considered in future testing programs.

Hydrogen has been colloquially referred to as a "great equalizer" in fatigue measurements, nullifying many unique aspects between materials. Conversely, of the various alloys documented in this report, the upper and lower bound fit to the data corresponded to a variability of 1.5 to 2 orders of magnitude crack growth rate at the initial, knee, and final data point locations. While hydrogen does appear to reduce some of the complexity of line pipe steels, significant variability remains, further compounded by the limited number of studies completed. This range of variability appears to be similar to the 1.5 orders of magnitude variability in fatigue crack growth rates of the pipe body in an inert environment (Figure 6). Given the limited data points and range of materials studied, it is reasonable to conclude that the upper and lower bounds shown in Figure 6 and Figure 8 might not completely encompass the observed variability in line pipe steels.

When considering the direction of testing for pipeline materials in hydrogen service, the observations presented in a foundation report from 1982 are likely still valid (Figure 11) and support the recommended future testing.

Results of Review by Pipeline Personnel

To help ensure that the research approach being used by Battelle is the most efficient method for achieving the long-range goals, the present report and Battelle's plan for future research was submitted to four pipeline experts from the natural gas transmission industry for their review. The comments of those experts is included in an appendix to this report. The principal points raised by the reviewers were:

(1) Suitability of newer, higher-strength pipeline steels for hydrogen service in regard to possible problems with subcritical crack growth

(2) How to define suitability of existing pipelines for conversion to hydrogen service.

(3) Concern for failure at hard spots and hard heat-affected zones in existing pipelines.

(4) More emphasis needed on long-term, full-scale tests

(5) Emphasis on actual double-submerged-arc seam and girth welds from actual pipe

(6) Concern for problems associated with hot-tapping

(7) Additional study of inhibiting effects of gas additives other than oxygen.

Battelle views the comments made by the pipeline experts as most helpful and is incorporating those aspects into the future work plans.

Figure 11 – Transmission Pipeline Expert Opinions on Initial Hydrogen Research by Battelle, Circa 1982 [37]

6. RECOMMENDATIONS TO SUPPORT FUTURE HYDROGEN PIPELINES

This report studied and compiled several decades of research on hydrogen's impact on materials commonly used in the natural gas transmission pipeline industry. The proposed hydrogen adsorption mechanisms were not the focus of this study but rather the impact of hydrogen on the mechanical properties of vintage and modern pipeline steels. Published studies from US government labs (e.g., National Institute of Standards and Technology (NIST), Sandia) and global publications were reviewed and summarized on API 5L and CSA Z245 steels. This project focused on the gathering, dissemination, and identification of critical variables affecting line pipe steels in hydrogen service. These factors include mechanical, microstructure, chemical properties, and steel making practices. Key variables significantly influencing fatigue performance and gaps in testing were identified, providing guidance to industry in future testing programs for the safe operation of hydrogen pipelines. The influence of hydrogen on line pipe steels could be summarized as:

- The use of low strain rates during tensile testing results in the greatest reduction in ductility. Hydrogen's impact on yield and ultimate tensile strength was not studied in detail as little or no impact was observed.
- Crack orientation significantly affected fatigue crack growth rate (FCGR), where higher FCGR was observed parallel to rolling direction. FCGR was reduced when the crack was oriented perpendicular to rolling direction.
- Banded ferrite-pearlite structures, common in vintage steels, provide improved FCGR resistance than ferrite/polygonal ferrite microstructures (modern, lower grade steels).
- Polygonal ferrite and alloys with higher fractions of polygonal ferrite have poorer fatigue performance. Materials with higher fractions of PF are often lower grade and associated with slower cooling and larger grain sizes.
- Rapid strain rate tests (e.g., Charpy V-notch (CVN)) should be avoided and may indicate that traditional CVN toughness conversions (CVN-to-CTOD or CVN-to-K) that are commonly utilized for the integrity management of vintage pipelines are likely unreliable for hydrogen pipelines and slow strain rate toughness measurements (e.g., Crack tip opening displacement (CTOD)) would be required.
- Quench and temper heat treatment conditions are more favorable than normalized conditions. This may indicate that normalized electrically-welded (EW)/electrical resistance weld (ERW) long seams, common in modern pipe, may exhibit higher FCGR.
- Girth weld and long seam heat affected zone (HAZ) experiences higher FCGR than weld metal or base metal whereas the base metal (pipe body) and weld metal generally exhibited similar FCGRs.
- The presence of very small imperfections significantly affected FCGR results.

A brief comparison of materials tested for application in hydrogen service to the current pipeline network, including vintages, grades, and existing pipeline damage, is also completed to identify future testing needs for the safe introduction of hydrogen into the pipeline network. Of the materials studied, only 22 unique materials were studied for their fatigue performance in hydrogen despite the hundreds of publications available. When compared to the existing pipeline network, where the largest fraction of installed pipe is pre-1970s vintage pipe, an insufficient variety of tested materials are available. The insufficient number of tests extended to the fatigue performance of girth welds and long seam welds. When comparing the tested materials to the existing pipeline network and modern pipeline materials, the following gaps are observed:

i. The large fraction of installed vintage pipe, which possess varying quality between mills (e.g., low-frequency electric resistance weld (LF-ERW) challenges) and higher likelihood of imperfections from manufacturing and decades of operation, necessitates significantly more testing on a greater variety of materials.

ii. All pipelines contain a significant number of welds, both girth weld and long seam, and an expansion in the number of studied welds is required. The majority of installed pipelines are girth welded with shielded metal arc welding (SMAW), a manual process that can produce significant variability from weld to weld. The variability within girth welds is further compounded with the intermittent girth weld inspections and presence of defects larger than those found in this report. The recent challenges with HAZ softening in modern high strength pipe should also be investigated where a decrease in yield strength results in strain concentration and additional hydrogen degradation.

A statistically significant expansion in the testing program to fully characterize and quantify the impact of hydrogen on pipelines is required.

DISCLAIMER

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Table 4 – Summary of Unique FCGR Studies.

Reference	Publication Year	Country	Specification & Grade	Manufacturing Year	Mill	Outside Diameter	Thickness	Seam Type	Chemistry Available	Microstructure Available	Test Medium	Fatigue Test Method	Test Pressure, max	H_2 Partial Pressure
						mm (NPS)	mm (in.)		Y/N	Y/N			MPa (psi)	MPa (psi)
[127]	2019	USA	API 5L Gr B	-	-	-	-	-	-	-	Gaseous	-	20.7 (3000)	20.7 (3000)
[52]	1985	USA	API 5L X42	-	-	323.9 (NPS12)	9.5 (0.375)	ERW	Y	Y	Gaseous	ASTM E647-83	6.9 (1000)	6.9 (1000)
[37]	1982	USA	API 5L X42			323.9 (NPS12)	9.5 (0.375)	-	Y	-	Gaseous	Full-Scale Fatigue	13.8 (2000)	13.8 (2000)
[24], [102]	2012	USA	API 5L X52	2000s	-	508 (NPS20)	12.7 (0.500)	-	Y	N	Gaseous	ASTM E647-08	-	-
[101], [97], [100], [103], [121]	2019	USA	API 5L X52	1964	-	914 (NPS36)	10.6 (0.417)	DSAW	Y	Y	Gaseous	ASTM E647-11	34 (4900)	34 (4900)
[101], [97], [100], [103], [121]	2019	USA	API 5L X52	2011	-	508 (NPS20)	10.6 (0.417)	ERW	Y	Y	Gaseous	ASTM E647-11	34 (4900)	34 (4900)
[87], [121]	2013	USA	API 5L X52	1990s	OSM Tubular, Camrose, CA	323.9 (NPS12)	12.7 (0.500)	ERW	Y	Y	Gaseous	ASMT E647-05	21 (3045)	21 (3045)
[116]	2018	USA	API 5L X60	1980	-	-	-	-	-	-	Gaseous	-	6.9 (1000)	6.9 (1000)
[122]	2021	France	API 5L X60	1972	-	350	7 (0.276)	-	Y	Y	Gaseous	-	8.5 (1230)	2.1 (305)
[40], [72]	2010	USA	API 5L X60	-	-	Skelp	-	-	Y	Y	Gaseous	ASTM E647-08	21 (3045)	21 (3045)
[57], [61], [121]	2016	USA	API 5L X65	-	-	508 (NPS20)	25.4 (1.00)	-	Y	Y	Gaseous	ASTM E647-08	21 (3045)	21 (3045)
[101], [97], [100], [103], [121]	2019	USA	API 5L X70	2005	-	914 (NPS36)	18 (0.709)	SAW	Y	Y	Gaseous	ASTM E647-11	34 (4900)	34 (4900)
[101], [97], [100], [103], [121]	2019	USA	API 5L X70	2005	-	914 (NPS36)	22 (0.866)	-	Y	Y	Gaseous	ASTM E647-11	34 (4900)	34 (4900)
[39]	2019	Norway	API 5L X70	-	-	-	23.3 (0.917)	-	Y	Y	Aqueous	ASTM E647	-	-
[34], [134]	2012	France	API 5L X80	-	-	914 (NPS36)	12.7 (0.500)	-	Y	N	Gaseous	ISO 12135	30 (4400)	30 (4400)
[40], [72]	2010	USA	API 5L X80	-	-	Skelp	-	-	Y	Y	Gaseous	ASTM E647-08	21 (3045)	21 (3045)
[135]	2017	China	API 5L X80	-	-	-	-	-	Y	N	Gaseous	ASTM E647	12 (1740)	6 (870)
[24], [102]	2013	USA	API 5L X100	-	-	1320 (NPS52)	20.6 (0.811)	-	Y	Y	Gaseous	ASTM E647-08	6.89 (1000)	6.89 (1000)
[60], [64]	2018	USA	API 5L X100	1990s	-	1300	19		Y	Y	Gaseous	ASTM E647-	21 (3045)	21 (3045)

Reference	Publication Year	Country	Specification & Grade	Manufacturing Year	Mill	Outside Diameter	Thickness	Seam Type	Chemistry Available	Microstructure Available	Test Medium	Fatigue Test Method	Test Pressure, max	H_2 Partial Pressure
							(0.750)					11		
[37]	1982	USA	A106 Gr. B	-	-	101.6 (NPS4)	6 (0.237)	SMLS	Y	-	Gaseous	ASMT E647-81	6.9 (1000)	6.9 (1000)
[38]	2016	Japan	JIS-SM490B	-	-	Plate	-	-	Y	N	Gaseous	ASTM E647-08	0.7 (100)	0.7 (100)
[136]	2017	USA	JIS-SS400	-	-	Plate	-	-	Y	Y	Gaseous	ASTM E647	40 (5800)	40 (5800)