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Application of Paris' law for estimation of hydrogen-assisted fatigue crack growth

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1 Abstract

Based on the general form of Paris' law a new method is proposed for estimating the rate of cracking of metals under Hydrogen-Assisted Fatigue. It is based on relating the fatigue crack growth rate of hydrogen embrittled metal to the fatigue crack growth rate of the metal without hydrogen embrittlement. One-dimensional hydrogen diffusion is assumed. Simulations using the proposed method on steel types X52, X70-80 and X80 are in agreement with published experimental tests results. The new method obviates numerical modelling of crack propagation and much reduces the computational costs.

2 Introduction

Hydrogen has a detrimental effect on metals [1, 2] through reducing ductility [3, 4] and thus increasing rate of fatigue cracking [5, 6]. Some stainless steels are particularly susceptible to hydrogen embrittlement when mechanically formed into shape for pipelines [7]. Hydrogen can be accessible to the steel in pipelines in several different ways. For example, pipes conveying hydrogen (as a fuel) are directly prone to the possible damaging effects of hydrogen. Despite this, steel remains the first choice for pipelines [8-10] in the expanding market for hydrogen fuel [9, 11]. Development in recent years of design codes for hydrogen piping and pipeline, such as ASTM B31.12 [12] is the direct result of the increasing demand for conveyance of hydrogen using steel pipelines. Another example is that hydrogen may contact the internal surface of pipes when a contaminant such as H_2S is part of the conveyed fluid, such as in sour gas pipelines [13, 14]. Further, aqueous soil environments can generate hydrogen electrochemically on the pipe surface [15-18]. All these highlight the necessity of reliable methods for evaluating hydrogen influence on metallic component such as pipelines and other components when hydrogen may be involved in some way in the transmission system.

Computational models for representing Hydrogen-Assisted Fatigue Cracking (HA-FC) usually are classified in two general categories according to the effect hydrogen has on material fracture properties [10]. In the first category the failure of a hydrogen affected region is attributed to "hydrogen enhanced decohesion" (HEDE). In this approach a so-called "traction separation law" [19-22] is used to overcome, in numerical modelling, the singularity at the crack tip. The second category includes methods in which failure is attributed to "hydrogen enhanced local plasticity" (HELP). Typically in both cases numerical modelling usually employs Finite Element methods. These are known to have high

computational requirements and high associated costs. The present study proposes a different approach, focussed on HA-FC. This approach does not involve finite element modelling and thus has considerably lower computational costs. The approach proposed herein is based on the use of Paris' law to estimate the rate of cracking under a HA-FC mechanism. The general form of Paris' law proposed in [23] is invoked and the crack growth rate of an embrittled metal is estimated by using the device of magnifying the growth rate that would have occurred in absence of hydrogen in the virgin metal. The overall basis for this approach is outlined in the next section. The following section outlines details of the modelling employed. In brief, the modelling approach proposed herein uses Fatigue Crack Growth Rate (FCGR) data of virgin steels to estimate the Hydrogen-Assisted Fatigue Crack Growth Rate (HA-FCGR) for that steel. HA-FCGR behaviour is then considered in terms of three parameters, namely: a lower bound and an upper bound the rate of reduction of material toughness and stress intensity factor range respective to these bounds. The final part of the paper provides examples and a discussion of the results.

3 Hydrogen concentration ahead a moving crack tip

In this section diffusion of hydrogen from a moving crack tip is considered. The aim is to estimate the hydrogen concentration ahead the crack tip. The hydrogen concentration so obtained is then used to evaluate the level of embrittlement ahead the moving crack in a HA-FC process.

Hydrogen diffusion is driven by the gradient of chemical potential [22]. Several factors influence the gradient, including the extent of local plasticity, the rate of plastic strain, the presence of hydrogen trapping sites and the local stress field [24-27]. Only one-dimensional diffusion is considered herein. Diffusion in the direction normal to the crack plane has been reported as not influential in the distribution of hydrogen concentration ahead of the crack tip and thus is not of major influence in governing diffusion ahead of the crack tip [22, 28, 29].

Simple form of one-dimensional diffusion is:

$$\frac{\partial u}{\partial t} = D \frac{\partial^2 u}{\partial x^2} \quad (1)$$

in which:

u : Relative concentration of hydrogen in the metal, defined as

$$\left(\frac{\text{Number of hydrogen particles in a unit volume of the metal}}{\text{Number of hydrogen particles in a unit volume of the environment}} \right)$$

t : Elapsed time, starting from formation of a new crack tip

D : Apparent diffusivity of hydrogen in the metal

x : Distance in the direction of the crack, from the crack tip to the point where the hydrogen concentration is being considered

For simplicity let the relative concentration of hydrogen at the tip of the crack, where the metal is in direct contact with hydrogen, be assumed to be unity irrespective of the hydrogen pressure in

the surrounding environment. Also, it is assumed that hydrogen concentration in the environment is constant. $u=1$ at the crack tip means that hydrogen concentration in the metal at the crack tip is equal to the concentration of hydrogen in the environment (the number of particles in a unit volume is the same in the metal and environment). This means the boundary conditions are assumed to move with the crack as it grows. Thus, to solve equation (1) under moving boundary conditions the origin of the coordinate system is reset to be at the crack tip for each incremental step. In consequence, the boundary conditions for each increment of crack growth are:

$$u_n(0, t) = 1 \quad (2)$$

$$\lim_{x \rightarrow \infty} u_n(x, t) = 0 \quad (3)$$

in which u_n is the relative concentration of hydrogen in the metal after the occurrence of the n th increment.

Considering the moving boundary condition problem to solve equation (1), the initial condition for the next increment is the solution of the previous increment. Recalling that the origin is set at the crack tip, the initial condition for n th increment is as follows:

$$u_n(x, 0) = u_{n-1}(x + x_n, t_n) \quad (4)$$

where x_n and t_n are the crack growth and the time between the $(n-1)$ th and n th increments, respectively. According to the solution of one-dimensional homogeneous boundary-value problems of diffusion in a semi-infinite region when $u_n(0, t) = 0$ and under initial condition (4) the solution for u_n will be[30]:

$$u_n(x, t) = \sqrt{\frac{D}{4\pi t}} \int_0^\infty u_{n-1}(x + x_n, t) \left(e^{\frac{-D(x-z)^2}{4t}} - e^{\frac{-D(x+z)^2}{4t}} \right) dz \quad (5)$$

Also, when $u_n(0, t) = 1$ and the initial condition is zero, u_n will be:

$$u_n(x, t) = \operatorname{erfc} \left(\sqrt{\frac{1}{4Dt}} x \right) \quad (6)$$

Thus when boundary conditions (2) and (3) and initial condition (4) are applied, due to linearity of (1), the total solution for $u_n(x, t)$ will be:

$$u_n(x, t) = \operatorname{erfc} \left(\sqrt{\frac{1}{4Dt}} x \right) + \sqrt{\frac{D}{4\pi t}} \int_0^\infty u_{n-1}(x + x_n, t) \left(e^{\frac{-D(x-z)^2}{4t}} - e^{\frac{-D(x+z)^2}{4t}} \right) dz \quad (7)$$

In the formulation in equation (7) it is assumed that crack tip jumps instantaneously in a cycle. In reality the crack tip might grow continuously or have a more gradual movement, somewhere between jumping and a continuous movement. Evidently, the formulation in equation (7), which is based on discontinuous movement (jump) of the crack tip will estimate a lower hydrogen

concentration than would be obtained by the solution for continuous movement of the crack tip. To support the following arguments about concentration of diffused hydrogen ahead the crack tip, let equation (7) be used for estimating hydrogen concentration. Figure 1 shows the hydrogen concentration ahead of the crack tip when the crack growth rate varies between 10^{-8} (m/cycle) and 10^{-6} (m/cycle). The crack growth rates of 10^{-8} (m/cycle) and 10^{-6} (m/cycle) are associated with the $\Delta K = 8$ (MPa.m^{1/2}) and $\Delta K = 18$ (MPa.m^{1/2}) in the HA-FCGR curve for X-70-80 grade steel, (Figure 3). The hydrogen diffusivity (D) is taken as 7.5×10^{-11} (m²/s) in Figure 1, adopted from [31], and loading frequency is taken as 1Hz in all diagrams. Thus the time between two subsequent jumps is one second.

Figure 1- Relative hydrogen concentration ahead the crack tip with different crack growth rates. Hydrogen diffusivity is 7.5×10^{-11} (m²/s). Loading frequency= 1Hz.

The trend lines in Figure 1 show that for a realistic range of crack growth rates the relative hydrogen concentration is close to unity. It follows that hydrogen concentration is quite close to that of the surrounding environment along the path followed by the crack tip in the metal. Figure 1 is obtained under a constant hydrogen diffusivity of 7.5×10^{-11} (m²/s). However, the local apparent diffusivity can be significantly greater under the influence of a metal fatigue process. For example, it has been shown that the apparent hydrogen diffusivity under fatigue is some five orders of magnitude greater than the standard hydrogen diffusivity for type 304 austenitic stainless steel exposed to 10 MPa of hydrogen pressure [32]. It follows that the real hydrogen concentration ahead the crack tip can be even closer to unity than that suggested by Figure 1. To support this argument and show the influence of increasing hydrogen diffusivity on the hydrogen concentration ahead the crack tip, a crack growth rate of 10^{-7} (m/cycle) under varying hydrogen diffusivity is considered in Figure 2. Hydrogen diffusivity, D varies between 2×10^{-11} (m²/s) and 100×10^{-11} (m²/s) in this case. The crack growth rate of 10^{-7} (m/cycle) is associated with the $\Delta K = 10$ (MPa.m^{1/2}) in HA-FCGR curve of X-70-80 steel grade (Figure 2). As before, loading frequency is 1Hz in all diagrams.

Figure 2- Relative hydrogen concentration ahead the crack tip under different hydrogen diffusivity when crack growth rate is 10^{-7} (m/cycle). Loading frequency= 1Hz.

It can be inferred from Figure 2 that hydrogen concentration ahead the crack tip is even greater than shown in Figure 1. Overall, in the range of distances of the order of the growth in crack length in a cycle, the value of the relative hydrogen concentration is almost unity, indicating that the material is almost uniformly embrittled. This matter is considered in more detail in the next section.

The method presented herein is a simple analytical model for diffusion of hydrogen ahead the crack tip. The aim is to show that hydrogen concentration is almost constant in the region that crack grows in one cycle and values presented in Figure 1 and Figure 2 are not used herein. As mentioned above, there are some parameters other than what are considered here in the diffusion of hydrogen ahead a crack tip. However, the ignored factors do not make a sudden change in the trend of the diagrams in Figure 1 and Figure 2 (see [22]). Figure 1 and Figure 2 show that the distance that crack tip develops in one cycle is so short that no significant change in the hydrogen concentration along the crack path happens (regardless of what is the exact form of the concentration diagram). Therefore, crack tip moves in an almost uniformly embrittled path.

4 Rate of cracking in a HAFC process

The mechanisms for hydrogen diffusion in metals have been extensively studied and documented [27, 33, 34]. However, the effects of hydrogen on the fracture strength of steels is not as clearly understood [10]. Some studies suggest that material properties such as the elastic modulus and the yield strength are affected by hydrogen [35]. Also, the concentration of hydrogen in the neighbourhood of a crack is considered to have a direct effect on the local material toughness [36].

As noted in the previous section, even under the highest hydrogen-assisted crack growth rates in the data sets considered herein (10^{-6} (m/cycle)) the hydrogen concentration is quite close to unity along the crack path. This means that under a HA-FC mechanism the crack tip can be considered as moving in a homogeneously embrittled material. Also, evidence from fractography shows that, in

the presence of hydrogen, there is little or no plastic deformation when the crack has a planar growth pattern [13, 14, 37]. Therefore it is reasonable to assume that the plastic region, if it occurs, remains close to the crack tip in a HA-FC mechanism. Further, since the relative hydrogen concentration is almost unity at the crack tip and through to the region ahead of the crack tip that is likely to become plastic, it can be inferred that the condition in a HA-FC situation is similar to crack propagation in a new (homogeneous and more brittle) material. It follows that the equations conventionally used to estimate the crack growth rate in homogeneous materials can be adopted also for application to HSA-FC case.

To derive the relation between HA-FCGR and FCGR it is necessary to determine the relationship between ΔK (range of stress intensity factor) in the virgin (unembrittled) and that in the embrittled material. It appears that for many steels the elastic modulus does not change by more than 2% as a result of hydrogen embrittlement [22, 38, 39]. Thus, within the context of the so-called Small-Scale Yielding (SSY) condition, the stress field around the crack tip is controlled predominantly by crack size and by external loading. It follows, therefore, that for a specific crack length and for given loading, the stress field of interest may be assumed to be the same, both for the embrittled and for the virgin metal. In the following section crack propagation in the embrittled region (which will be considered as a new material) is discussed.

To make progress it will be assumed in the following that the relation between fracture toughness and hydrogen concentration can be represented by a linear function. Earlier, in [22], based on studies in [40, 41] a quadratic relation was proposed between hydrogen concentration and the reduction in cohesive energy. However, for the range of hydrogen concentration possible in realistic situations a linear relation between hydrogen concentration and reduction in cohesive energy is a close approximation. It was used also in earlier studies [29, 42, 43]. Assuming linear elastic fracture mechanics (LEFM) provides a valid representation of the fatigue cracking problem, it follows that the cohesive energy is equal to the critical energy release rate, which in turn has a linear relationship with fracture toughness [44]. As a result it is assumed here that a linear relation exists between fracture toughness, K_{IC} , and hydrogen concentration, u :

$$K_{IC,H} = K_{IC}(1 - \mu \times u) \quad (8)$$

$K_{IC,H}$: Fracture toughness of the uniformly embrittled metal

K_{IC} : Fracture toughness of the virgin metal (i.e. not embrittled)

u : Hydrogen concentration in the metal

μ : Reduction factor for fracture toughness reduction caused by hydrogen concentration

The assumption of linear relation between hydrogen concentration and fracture toughness is made just for the region ahead the crack tip in which crack grows. This is highly localised region ahead the crack tip. Based on the results shown in Figure 1 and Figure 2 hydrogen concentration is almost constant in the range of crack growth in one loading cycle.

No significant change of yield stress under hydrogen embrittlement has been reported for the steel grades modelled in the present study [45]. Equation (9) shows this relation:

$$\sigma_{y,H} = \sigma_y \quad (9)$$

in which:

$\sigma_{y,H}$: Yield stress of uniformly embrittled metal

σ_y : Yield stress of the virgin metal (not embrittled)

The above discussion has provided understanding about the basic assumptions in the proposed model. To develop a formulation between rate of cracking and range of stress intensity factor general form of Paris' law is conserved. In [23] based on dimensional analysis and incomplete self-similarity, Paris' law [46] is formulated as follows:

$$\frac{da}{dN} = \frac{\Delta K^{2+\alpha(R,Z)}}{\sigma_y^2 K_{Ic}^{\alpha(R,Z)}} \Phi(R, Z) \quad (10)$$

in which:

$$Z = \frac{\sigma_y \sqrt{h}}{K_{Ic}}$$

R : Load ratio, defined as the ratio of the minimum and maximum loads in a cycle

h : A characteristic specimen length scale

$\frac{da}{dN}$: Crack growth rate

ΔK : Range of stress intensity factor in a loading cycle

Since the Paris' law applies to homogeneous materials, it follows that equation (10) applies when material is homogeneously embrittled. It then follows that equation (10) can be used also in the case of HA-FC material, based on the argument presented above regarding homogeneous embrittlement of material around the crack tip.

To derive a relationship between Hydrogen-Assisted Fatigue Crack Growth Rate ($\frac{da_H}{dN}$) and the range of stress intensity factor (ΔK), use can be made of equation (8) to equation (10). Two cases can be considered: fatigue cracking in the absence of hydrogen and fatigue cracking assisted by hydrogen. The crack length and the loading are assumed to be the same for both cases. As mentioned above, ΔK will be the same when the SSY (Small Scale Yielding) condition is valid. With these assumptions, using equation (8) and (9), and with a specific value of ΔK , the corresponding crack growth rate under hydrogen assisted fatigue $\frac{da_H}{dN}$ will be:

$$\frac{da_H}{dN} = \frac{\Delta K^{2+\alpha_H(R,Z)}}{\sigma_y^2 K_{Ic}^{\alpha_H(R,Z)} (1 - \mu)^{\alpha_H(R,Z)}} \Phi_H(R, Z) \quad (11)$$

Attention is turned now to μ in equations (8) and (11). In equation (8), μ indicates the influence of hydrogen concentration on the decrease in material toughness. For a limited range of change in stress intensity factor μ is assumed to be constant [29]. In the model proposed herein a lower bound, ΔK_0 , and an upper bound, ΔK_1 , are introduced to bound the stress intensity factor, corresponding to the range of values in μ . Also, a maximum value for μ is adopted, termed μ_{max} . Further, let ΔK_1 be the value of ΔK at which μ is essentially insensitive to ΔK . Physically it indicates that although in general μ increases by increasing ΔK , there is an upper limit to this phenomenon. In the same sense a lower limit for influence of ΔK is considered herein below which μ is assumed not affected by ΔK . Let this lower limit of ΔK be denoted ΔK_0 .

Following on from the above assumptions, it is assumed that when ΔK is lower than ΔK_0 or when ΔK is greater than ΔK_1 , μ is constant. Further, when μ is constant, $2 + \alpha_H$ is the slope of the Fatigue Growth Rate curve in log-log coordinates, based on equation (11). $\alpha_H(R, Z)$ is a function of R and Z which are assumed not to change with embrittlement (as proposed above). Thus in the ranges of ΔK that μ is constant (μ is μ_{min} when $\Delta K < \Delta K_0$ and μ is μ_{max} when $\Delta K > \Delta K_1$) a constant slope is expected in the Fatigue Growth Rate curve on a log-log plot. This proposition is directly supported by experimental observations for HA-FCGR, as shown, for example, by curves such as shown in Figure 3. It shows that the HA-FCGR curve is quite close to a straight line in the regions marked "A" and "C". This suggests immediately that when $\Delta K < \Delta K_0$ or $\Delta K > \Delta K_1$, α_H can be estimated by finding the slope of HA-FCGR curve in a log-log coordinate system. Also, the value of ΔK for which the slope of HA-FCGR curve is constant for any lower value of ΔK then gives the lower bound, ΔK_0 . Similarly, the value of ΔK at which the slope of HA-FCGR curve is constant for any greater ΔK gives the upper bound, ΔK_1 . In this way ΔK_0 and ΔK_1 can be estimated by distinguishing regions "A" and "C" on the HA-FCGR curves.

Table 1 summarises the values of α_H , ΔK_0 and ΔK_1 calculated for each of three steel grades. The values of α_H (columns 6 and 8) in this table are compared with those of the virgin material (column 2). In Table 1 α_H of regions "A" and "C" are denoted by $\alpha_{H,A}$ and $\alpha_{H,C}$, respectively. $\alpha_{H,A}$ and $\alpha_{H,C}$ are obtained by fitting curves to the data points in regions "A" and "C" of the HA-FCGR curves. α (column 2) in this table is estimated by fitting curve to FCGR curves.

Figure 3- Hydrogen-Assisted Fatigue Crack Growth Rate curve of steel grade X70-80, Hydrogen pressure= 21MPa. Experimental data from [47].

Table 1- Values of α , ΔK_0 , ΔK_1 , $\alpha_{H,A}$ and $\alpha_{H,B}$ for three grades of steel under different hydrogen pressure								
Clmn1	Clmn2	Clmn3	Clmn4	Clmn5	Clmn6	Clmn7	Clmn8	Clmn9
Steel grade	α	Hydrogen Pressure (MPa)	ΔK_0 (MPa $\times$ m ^{1/2})	ΔK_1 (MPa $\times$ m ^{1/2})	$\alpha_{H,A}$	$100 \frac{ \alpha_{H,A}-\alpha }{\alpha}$ (%)	$\alpha_{H,C}$	$100 \frac{ \alpha_{H,C}-\alpha }{\alpha}$ (%)
X-80*	1.51	5.5	9.0	12.0	1.38	8.6	1.40	7.3
		21	8.2	12.0	1.37	9.3	1.60	6.0
X-70-80**	1.81	5.5	9.5	12.0	1.67	7.7	1.75	3.3
		21	8.3	12.0	1.85	2.2	1.98	9.4
X-52*	1.97	21	8.3	12.0	2.11	7.1	2.05	4.1

* [48], ** [47]

The exponent in Paris' law, $(2 + \alpha)$ in equation (10), is known to increase considerably in value as the material becomes more brittle [49, 50]. However, as shown in Table 1 the values of $\alpha_{H,A}$ and $\alpha_{H,C}$ are ultimately only some 9.4% different from the values for α for the data sets analysed herein. This indicates that α , a material property in equation (10), does not change very much as a result of hydrogen embrittlement and that the difference in the slope of curves for the HA-FCGR and FCGR cases is the result of changes in the value of μ with ΔK . In view of this observation it is assumed in the following that α_H in equation (10) can be taken as equal to α . This is based on the results presented in Table 1. Also, it has been shown previously that the value of Φ changes with changes in value of α [23, 51]. Taking both aspects into consideration suggests it is reasonable to assume that Φ_H is equal to Φ . Thus equation (11) becomes:

$$\frac{da_H}{dN} = \frac{\Delta K^{2+\alpha}}{\sigma_y^2 K_{Ic}^\alpha} \Phi \times [\lambda(\Delta K)]^\alpha \quad (12)$$

In which

$$\lambda(\Delta K) = \frac{1}{1 - \mu(\Delta K)}$$

Since $0 < \mu_{min} < \mu(\Delta K) < \mu_{max} < 1$, it follows that $1 < \frac{1}{1-\mu_{min}} < \lambda(\Delta K) < \frac{1}{1-\mu_{max}} < \infty$. Further, in the following let it be assumed that there is a linear relationship between $\lambda(\Delta K)$ and ΔK when $\mu_{min} < \mu(\Delta K) < \mu_{max}$ leading to:

$$\lambda(\Delta K) = \begin{cases} \lambda_{min} & \Delta K \leq \Delta K_0 \\ \lambda_{min} + \frac{\lambda_{max} - \lambda_{min}}{\Delta K_1 - \Delta K_0} (\Delta K - \Delta K_0) & \Delta K_0 < \Delta K \leq \Delta K_1 \\ \lambda_{max} & \Delta K_1 < \Delta K \end{cases} \quad (13)$$

in which:

$$\lambda_{min} = \frac{1}{1-\mu_{min}} \text{ and } \lambda_{max} = \frac{1}{1-\mu_{max}}$$

The relation between $\lambda(\Delta K)$ and ΔK is dependent on ΔK_0 , ΔK_1 , μ_{min} and μ_{max} . As mentioned above, the values of ΔK_0 and ΔK_1 can be estimated by distinguishing the regions "A" and "C" on the HA-FCGR diagram. Also, the parameters α and Φ for the Paris' law (equation (12)) may be obtained by fitting a curve to the FCGR data points. Herein let the values for μ_{min} and μ_{max} be estimated by fitting a curve to the experimental HA-FCGR data points. It is noted that μ_{min} has most influence on

the lower values of ΔK , i.e. those at the left of the simulated curve, as shown in Figure 4. It illustrates the four simulated HA-FCGR curves for different values of μ_{min} as shown.

Figure 4- Sensitivity of simulated results to μ_{min} . Steel grade: X70-80, Hydroge pressure= 21MPa. $\mu_{max} = 0.82$. Experimental results from [47].

The right-hand part of the simulated curve is influenced by μ_{max} . Figure 5 shows how the HA-FCGR curve changes with μ_{max} .

Figure 5- Sensitivity of simulated results to μ_{max} . Steel grade: X70-80, Hydroge pressure= 21MPa. $\mu_{min} = 0.55$. Experimental results from [47].

Once the influence of parameters μ_{max} and μ_{min} on the simulated HA-FCGR curve is known, the most appropriate value for each parameter can be estimated using a trial and error technique. It is noted also that each model parameter has a unique effect on the simulated HA-FCGR curve. Thus only one parameter is considered in the trial and error process. Table 2 shows the values obtained using this technique for μ_{max} and μ_{min} for three steel grades.

Table 2- Values of μ_{max} and μ_{min} for three grades of steel under different hydrogen pressure

Steel grade	Hydrogen Pressure (MPa)	μ_{min}	μ_{max}
X-80*	5.5	0.37	0.86
	21	0.53	0.86
X-70-80**	5.5	0.43	0.81
	21	0.55	0.82
X-52*	21	0.50	0.80

* adopted from [48], ** adopted from [47]

Using the values for α , ΔK_0 , ΔK_1 , μ_{min} and μ_{max} tabulated in Table 1 and Table 2, the results shown in Figure 6 and Figure 7 are obtained. Each shows both the experimental results, for X80 and X80-70 steel grades, extracted from the literature [47, 48] and the results obtained using the simulation method outlined above.

Figure 6- Simulated fatigue crack growth rate curves for X80 steel compared to experimental data from the literature. Right: hydrogen pressure= 21MPa, Left: hydrogen pressure = 5.5MPa (Parameters presented in Table 1 and Table 2). Experimental results adopted from [48]

Figure 7- Simulated fatigue crack growth rate curves for X80-70 steel compared to experimental data from the literature. Right: hydrogen pressure= 21MPa, Left: hydrogen pressure = 5.5MPa (Parameters presented in Table 1 and Table 2). Experimental results from [47].

5 Discussion

The results of the analyses given herein suggest that, μ (Reduction factor for fracture toughness reduction caused by hydrogen concentration) is influenced by ΔK in a Hydrogen-Assisted Fatigue Cracking mechanism. This phenomenon can be attributed to more accumulation of damage under higher ΔK values. This may be compared with damage models for cyclic loads. These consider that damage accumulates as a result of the occurrence of plastic strain during each loading cycle [52]. From this perspective damage at a point is determined by two factors, the strain magnitude and the number of cycles, both as they occur at that point. By comparing the exponent in Paris' law in different materials it can be inferred that this parameter increases with brittleness. Thus it can be expected that the exponent in Paris' law, $(2 + \alpha)$ in equation (10), will increase with hydrogen embrittlement. However, as Table 1 shows $\alpha_{H,A}$ and $\alpha_{H,C}$ ultimately differ only some 9.4% from α for the cases of data sets analysed herein. It suggests that instead of α , μ is the main cause of the difference in the slopes of the HA-FCGR and the FCGR curves of a specific metal. This may be explained as follows. Initially, consider the case in which we ignore μ from the calculations. Then fit a line to the HA-FCGR data points in log-log coordination system, consistent with the Paris' law ($\frac{da}{dN} = c\Delta K^m$). This situation is shown in Figure 8. By ignoring μ the exponent obtained from the Paris' law for the HA-FCGR curve will be greater than that for the FCGR curve (5.74 for the HA-FCGR curve and 3.58 for the FCGR curve). This result, that the exponent is greater for HA-FCGR than for FCGR, is in agreement with experimental observations (such as those illustrated in Figure 6 and Figure 7) reporting higher exponent for more brittle materials. However, as Figure 8 illustrates, using the simple form of Paris' law ($\frac{da}{dN} = c\Delta K^m$) results in considerable discrepancies between the fitted line and the experimental data in the HA-FCGR curve. On the other hand, the fitted line to the FCGR data in Figure 8 well represents the experimental results. It follows that in the absence of hydrogen, Paris' law provides an appropriate estimate of crack growth rate of the fatigue cracking. Nevertheless, when hydrogen diffuses from the crack tip into the metal the simple form of Paris' law is unlikely to be applicable and a simplified relation such as equation (12) can be used to estimate crack growth rate.

Figure 8- Experimental fatigue crack growth rate data for X80 and fitted curve when a simple form of Paris' law ($\frac{da}{dN} = c\Delta K^m$) is employed. Hydrogen pressure 21Mpa. Experimental data from [48].

For the modelled steel grades and based on the results presented in Table 1 (column 5) ΔK_1 is insensitive to both hydrogen pressure and the grade of steel, ΔK_0 , however, decreases with hydrogen pressure. It is, however, possible that both ΔK_1 and ΔK_0 are sensitive to microstructure of the steel and/or gas pressure. Equation (7) shows the hydrogen concentration in the metal relative to the surrounding environment. Therefore, under a specific relative hydrogen concentration in the metal by increasing the hydrogen pressure in the environment the overall hydrogen concentration in the metal increases. Thus a decrease of ΔK_0 under higher hydrogen pressure can be attributed to an increasing presence of hydrogen molecules in the metal when the hydrogen pressure increases. Further, it is noted that ΔK_0 changes with the grade of steel, as shown in Table 1. Thus the value of ΔK_0 range of stress intensity factor from which the apparent hydrogen diffusivity starts increasing, is likely to depend both on hydrogen pressure and on the metallurgical properties of the steel.

When ΔK is less than ΔK_{min} , μ equals μ_{min} . Therefore, for the case of a constant loading (when $\Delta K = 0$) μ is expected to be μ_{min} . Results of the experimental tests on X-100 are in agreement with this expectation. For steel grade X-100 under a constant loading μ is measured at 0.2 [29]. This value of μ (0.2) is close to the estimated values for μ_{min} for X-80 in Table 2-column 4.

Herein it is shown that when $\Delta K_0 < \Delta K < \Delta K_1$ a linear relation between $\frac{1}{1-\mu}$ and ΔK results in an appropriate fit to the experimental data. The parameter $\frac{1}{1-\mu}$ (which is termed λ) appears in the equation (12). This relationship appears to provide reasonable estimates of crack growth rate but the actual physical interpretation of relationships such as the linear relationship between $\frac{1}{1-\mu}$ and ΔK could be the subject of further investigation.

Finally, the proposed model is tested for three grades of steel and under two levels of hydrogen pressure. For these steel grades a linear relation between λ and ΔK well fits the simulated HA-FCGR curve to the experimental results. Applicability of linear relationship between λ and ΔK should be investigated for other metals. Also, it should be highlighted that the proposed model requires calibration of the parameters μ_{min} and μ_{mam} for every hydrogen concentration/pressure in the environment.

6 Conclusion

The experimental data sets analysed herein indicate that when hydrogen embrittlement occurs the simple form of Paris' law is unlikely to be appropriate to estimate the fatigue crack growth rate. A new relation is derived herein, developed from the general form of Paris' law [23]. The proposed relation results in a better fit to experimental data, at least for the data sets analysed. The parameters of the proposed model change according to the grade of steel, indicating that these parameters are related to the metallurgical properties of the steels considered.

It is also proposed that the Rate of reduction of material toughness (μ) can be incorporated in the relationship between crack growth rate ($\frac{da}{dN}$) and the (range of) Stress Intensity Factor (ΔK). The results indicate that the parameters in Paris' law are reasonably insensitive to hydrogen embrittlement. This implies that the difference between the HA-FCGR and the FCGR curves is the

result of the change in value of μ with ΔK . Further, only a limited range of Stress Intensity Factor (ΔK) was found to have a significant effect on μ .

The lower bound of effective range of ΔK on μ , ΔK_0 decreases with increase of hydrogen pressure for three of the steel types considered. However, the upper bound of the effective range of Stress Intensity Factor on μ is insensitive both to hydrogen pressure in the environment and to the grades of steel considered.

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