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EFFECT OF HYDROGEN ON THE BENEFIT OF WARM PRESTRESSING

For: SINTAP

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Prepared for: SINTAP

Contact: Dr R A Ainsworth
Nuclear Electric Ltd

Prepared by: Dr A Muhammed
Structural Integrity Department

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NOMENCLATURE

CTOD	Crack tip opening displacement
K	Stress intensity factor
K_c	Baseline fracture toughness in the absence of warm prestressing
K_f	Predicted final fracture toughness after warm prestressing
K_{mat}	Final fracture toughness after warm prestressing
K_1	Pre-load K
SENB	Single edge notch bend specimen
WPS	Warm prestressing
σ_{Y1}	Yield strength at pre-load temperature
σ_{Y2}	Yield strength at final fracture temperature

CONTENTS

EXECUTIVE SUMMARY

Background

The beneficial effect of warm prestressing (WPS) on subsequent fracture behaviour has been extensively investigated. However, much of the investigations do not consider the effect of in-service degradation such as hydrogen embrittlement. It is known that hydrogen has the effect of reducing fracture resistance, but the interaction of this effect with warm prestressing is not clear. The primary objective of this work is to determine the effect of hydrogen embrittlement on the benefit of warm prestressing.

Work Undertaken

Fracture toughness tests were carried out on specimens prepared from A533B pressure vessel steel. Some of these were tested in the as-received condition, while others were tested after hydrogen charging or proof loading or both. The tests conducted without proof loading were used to establish the baseline toughness, which was then compared with the corresponding results after proof loading. The general trends of experimental results were established. Also, the measured toughness values were compared with those obtained from theoretical WPS predictive models.

Main Conclusions

The following conclusions were drawn from comprehensive analyses of the test results:

- Hydrogen charging reduced the baseline fracture toughness of the material. The CTOD was on average reduced by 35% while the reduction in K was smaller at about 9%, for tests carried out at -100°C.
- Significant benefit of warm prestressing was observed in both uncharged and charged specimens at -170°C. The results suggest that most of the WPS benefit (up to 90% of K_{mat} in these tests) is retained after hydrogen charging. However, the proof load levels used in the tests carried out at -100°C were too low to give any significant residual benefit after hydrogen charging.
- The accuracy of WPS predictive models appears to be slightly affected by hydrogen charging. In general, the level of conservatism in the predictions were less for charged specimens than for uncharged ones. However, the predictions for both uncharged and hydrogen-charged specimens were on average conservative.
- Overall, the results suggest that existing theoretical models may be used to predict fracture toughness even after exposure to hydrogen. However, it should be recognised that the margins on these predictions may be slightly reduced compared to the cases without hydrogen effects. Using reduced baseline toughness values and the appropriate yield properties for the hydrogen-embrittled state should maintain good safety margins in the predictions.

1. INTRODUCTION

The beneficial effect of warm prestressing (WPS) on subsequent fracture behaviour has been extensively investigated. However, much of the investigations do not consider the effect of in-service degradation such as hydrogen embrittlement. It is known that hydrogen has the effect of reducing fracture resistance, but the interaction of this effect with warm prestressing is not clear. The primary objective of this work is to determine the effect of hydrogen embrittlement on the benefit of warm prestressing.

Fracture toughness tests were carried out on specimens prepared from A533B pressure vessel steel. Some of these were tested in the as-received condition, while others were tested after hydrogen charging or proof loading or both. The tests conducted without proof loading were used to establish the baseline toughness, which was then compared with the corresponding results after proof loading. The general trends of experimental results were established. Also, the measured toughness values were compared with those obtained from theoretical WPS predictive models.

2. EXPERIMENTAL PROGRAMME

2.1. GENERAL

The test programme was carried out in two phases. In the first, a total of 25 Bx2B single edge notch bend (SENB) specimens were tested. The specimens were 22mm x 44mm in cross-section, extracted parallel to the rolling direction and notched through the plate thickness in the L-T orientation. They were divided into five sets of five specimens. Two sets were used as control sets and were tested without proof loading, one in the as-received condition, the other after hydrogen-charging. These were aimed at establishing typical baseline toughness values in the absence of proof loading (with and without hydrogen). Of the remaining three sets, one was fractured after proof loading while two were fractured after both proof loading and charging. The two sets of specimens receiving both proof loading and hydrogen-charging had the sequence of loading and charging reversed to see if the sequence had a significant effect.

The first 25 specimens (M02) were tested at -100°C. This temperature was selected as a compromise temperature at which the effect of hydrogen and warm prestressing (WPS) were both expected to be significant. It is well known that the largest WPS benefits are obtained on the lower shelf where cleavage behaviour dominates (e.g., see Ref.1-3). On the other hand, the toughness degradation effect of hydrogen is maximum at ambient and diminishes with decreasing or increasing temperatures (see Ref 4). However, on completion of the first phase of testing, a significant number of specimens did not fail by cleavage. Some specimens failed after significant stable crack extension while others reached maximum load without fracture. In order to investigate the effect of hydrogen when the full benefit of WPS is applicable, it was decided to reduce the test temperature to -170°C in the second phase of experiments to ensure lower shelf behaviour in all specimens. A total of 15

22mm x 44mm SENB specimens (M03) were tested in the second phase of experiments.

2.2. MATERIALS

The material selected for the experiments was an A533B Class 1 steel. This material has been used extensively on previous TWI projects (e.g., Ref.1) and has been well characterised. The chemical analysis of the steel is reproduced in Table 1. The microstructure is tempered bainite with ASTM grain size between 7 and 8. Previous mechanical tests also gave the average yield strength of 506MPa and a tensile strength of 651MPa at room temperature. The yield strength at the final test temperature was estimated as 639MPa and 858MPa at -100°C and -170°C respectively. Pellini drop weight tests gave a nil-ductility transition temperature of -50°C.

Other data available from previous TWI prior overload projects on A533B material include Charpy and fracture toughness transition curves for different specimen and notch orientations. The effect of different levels of proof loading on fracture toughness (in terms of K and CTOD) was also investigated in Ref.1. This wealth of existing data provided a good resource for broad comparison with the results of the present study.

2.3. SPECIMEN PREPARATION AND HYDROGEN CHARGING

2.3.1. Tests at -100°C

The fracture toughness test pieces (B=22mm x 2B=44mm SENB) were taken from nominally 50mm thick parent plate (TWI log Ref.2A195) and notched through the thickness, in the L-T orientation. The specimens were fatigue pre-cracked from the machined notch in accordance with the procedures given in BS 7448:Part1:1991, to an initial crack depth ratio a/W=0.5. The specimens were given TWI identification numbers M02-1 to M02-25, that is, parent material, M02 specimens 1-25. The specimens were divided into five sets as follows (Set A = 1-5, Set B = 6 – 10, Set C = 11-15, Set D = 16-20 and Set E = 21-25)

Three sets (B,C,D) of five specimens needed hydrogen charging prior to final fracture. Corrosive charging was adopted for these specimens. The solution used was 5%NaCl, 0.5% acetic acid in water saturated with H_2S. The use of H_2S is appropriate as it may be encountered in a range of service and it is known to promote hydrogen ingress into the steel. The specimens were immersed in the solution at room temperature for 96 hours to ensure saturation with hydrogen. After the charging period, the specimens were stored in solid carbon dioxide until final fracture tests were carried out. The specimens were thus maintained at low temperature to prevent hydrogen loss prior to testing.

Specimen sets B and C were charged straight after machining and fatigue precracking, while set D was proof loaded before charging. In all cases, the notch tip was sealed with silicon rubber to prevent exposure of the crack tip to the solution. It was felt that direct exposure could lead to corrosion and possibly crack blunting. This could in turn cause higher fracture toughness measurements that were not directly attributable to the pre-load.

2.3.2. Tests at -170°C

Under the second phase of experiments, a total of 15 22mm x 44mm SENB specimens were tested. The specimens, given the TWI identification M03-1 to M03-15, were prepared largely as described above in Section 2.3.1 for the first phase. They were tested in 3 sets of five specimens. One set (F) in the as-received condition, the second (set H) after proof loading and the final (set G) after proof loading and hydrogen charging.

2.4. SPECIMEN PRELOADING

Apart from the control specimens, all specimens were preloaded at room temperature prior to final fracture. Each SENB specimen was instrumented with a double clip gauge arrangement across the notch mouth. Testing was carried out in a universal testing machine. The level of proof loading was carefully set to avoid excessive plasticity ahead of the crack tip. For the first few specimens, the load-clip gauge trace was monitored during loading and stopped shortly after the trace began to deviate from a linear-elastic regime (see Fig.1). This loading termination point was approximately achieved at a crack mouth opening displacement of about V_g = 0.5mm in all specimens. This did not give a constant amount of pre-load (or K) because of the inevitable variation in actual crack depth (a/W) between specimens.

The pre-load K, CTOD and J were calculated at the maximum load applied during the proof loading using the equations given in BS 7448:1991. The variation in the level of pre-load K was approximately ±4% about an average value of $3011Nm^{-3/2}$. The specimens were fully unloaded after reaching the fixed displacement (V_g =0.5mm) at room temperature. The loading and unloading was done at a fixed displacement rate of 0.50mm/min. The specimens were ultimately (some after hydrogen charging) cooled to -100°C or -170°C and then fractured. The entire testing cycle is commonly referred to as Load, Unload, Cool, Fracture (LUCF).

2.5. FRACTURE TOUGHNESS TESTING

As for the preloading, the double clip gauge arrangement was adopted for the final fracture loading. The testing was broadly in line with BS 7448:1991. The uncharged specimens were tested at a fixed displacement rate of 5.0mm/min while the charged specimens were tested at a displacement rate of 0.01mm/min. Based on previous experience at TWI[5], this slower strain rate was selected to ensure detection of hydrogen embrittlement effects. Loading was steadily increased until fracture (with or without tearing) or the attainment of maximum load. Again, the fracture parameters K, CTOD and J were calculated at the fracture or maximum load.

Details of the proof loading and final fracture testing are given in Appendix I. The results of the proof loading are identified by the letter P. For example P03 01 refers to the proof loading of specimen M03 01.

2.6. HYDROGEN ANALYSIS

In the first phase of experiments, two specimens were selected from each set (B, C and D) that was charged prior to fracture testing. After testing, samples typically 22mm x 44mm section by 10mm thick were carefully cut from the selected samples for hydrogen analysis. The samples were taken in the region immediately adjacent to the notch in order to get an indication of hydrogen levels in the crack tip region during the fracture test. The amount of diffusable hydrogen evolved over mercury was measured. This procedure was repeated for all five charged specimens tested in the second phase.

Details of the hydrogen charging and analysis are in Appendix II.

3. ANALYSIS OF TEST DATA

The fracture toughness parameters generated from the tests include K, CTOD and J values both at proof loading and at fracture. These were used to produce scatter plots to see if any obvious trends could be observed. For the second phase of experiments at -170°C, the final fractures were by cleavage and the pre-conditions for applying theoretical WPS predictive models were met. The measured values for both charged and uncharged specimens were therefore compared to theoretical predictions. This exercise was aimed at investigating if the level of conservatism in the models was influenced by the presence of hydrogen.

The two theoretical models used in this study, both of which applies to the LUCF cycle, have been widely published (see Ref.1 and 2). The Smith's model gives the failure stress intensity factor K_f at temperature T_2 after an applied preload stress intensity factor K_1 at temperature T_1 as:

$$\frac{K_f}{K_C} = \left(\frac{\sigma_{Y2}}{\sigma_{Y1} + \sigma_{Y2}}\right)\left(1 + \frac{\sigma_{Y1}}{\sigma_{Y2}}\frac{K_1}{K_C}\right) \qquad [1]$$

Where in the above equation, K_C is the fracture toughness at T_2 in the absence of a prestress; σ_{Y1} and σ_{Y2} are the yield strengths at T_1 and T_2, respectively. The second model used is that by Chell and Haigh[3] and takes the form:

$$\frac{K_f}{K_C} = 0.20\frac{K_1}{K_C} + 0.87 \qquad [2]$$

The terms have the same meaning as in Eq.[1].

4. RESULTS AND DISCUSSION

4.1. HYDROGEN CONTENT

Hydrogen analysis was carried out soon after fracture testing except for four specimens that had to be stored for two weeks over the end of year holiday period. The amount of hydrogen measured in the specimens was in the range 1.08 to 1.74ml/100g, except for those stored cold for two weeks prior to hydrogen analysis. It is clear that these specimens lost some hydrogen with measured levels in the range 0.14 to 0.73ml/100g. As all 15 charged specimens were charged in exactly the same way, there is no reason to believe that these specimens had significantly lower hydrogen content than the rest. Ignoring the samples that were in long term storage, the average hydrogen content from the samples was about 1.50ml/100g. Previous experience[5,6] has shown that the effect of hydrogen on fracture toughness, saturates at a concentration of about 1.0ml/100g. Therefore, it would be expected that the results obtained in this study would not be significantly altered by using higher levels of hydrogen concentration.

4.2. BASE FRACTURE TOUGHNESS AND HYDROGEN EFFECT

The results of the final fracture tests are given in Table 2 and 3. The results are also presented in the scatter plots of Fig.2-5. For the tests at -100°C, it can be seen from Table 2 that most of the specimens failed by cleavage (some after ductile tearing). However, three of the specimens reached maximum load after which the load began to drop and were subsequently unloaded. However, at the lower temperature of -170°C, lower shelf behaviour is observed with cleavage failure in all specimens.

The baseline CTOD values measured in non-proof loaded and uncharged specimens broadly agree with those reported for A533B material in a previous TWI project[1]. As would be expected, the values found here in 22mm thick specimens are slightly higher than those measured in 50mm thick specimens in Ref.1.

Figures 2 and 3 show that the baseline fracture toughness at -100°C is moderately reduced by hydrogen charging. From Table 2, the average measured base toughness without hydrogen (set A) is calculated in terms of K as $3751 Nmm^{-3/2}$ and CTOD as 0.17mm. The corresponding values after hydrogen charging (set B) are K = $3420 Nmm^{-3/2}$ and CTOD = 0.11mm. This is an average reduction in baseline toughness of 35% in CTOD and almost 9% in K as a result of hydrogen charging. This observation is in line with the findings of Humphries et al[5] on the study of hydrogen effects on A516 and A285 steels. In this reference, a major effect was found on CTOD with marginal reduction in K from tests carried out at ambient temperature. It is however noted that, the huge scatter in the baseline CTOD values observed in the present work may mean that the reduction of 35% is not typical.

The effect of warm prestress (WPS) on uncharged specimens is evident from Fig.2-5. From Table 2, the average increase in toughness at -100°C is 25% and 6% in terms of CTOD and K respectively. For the tests at -170°C, an average increase of 70% was observed in terms of K while the CTOD values after proof loading were on average almost three times the baseline value (see Table 3). As would be expected, higher WPS benefits were observed at -170°C where all tests were clearly cleavage dominated. Also, these tests involved higher pre-load K levels (as ratio of base toughness) than those at -100°C.

4.3. COMBINED EFFECT OF WARM PRESTRESS AND HYDROGEN

The effect of warm prestressing together with hydrogen at -100°C is not very clear. The samples receiving both warm prestress and hydrogen charging (sets C and D) do not give significantly different results to those hydrogen charged without any prestress (see Fig.2 and 3). It is also noted from Fig.2 and 3 that the sequence of proof loading and charging does not have a significant effect. However, the combined WPS and hydrogen effect is clearer for the tests at -170°C. Here, it is noted that the fracture toughness of preloaded charged specimens were significantly higher than the baseline toughness (see Fig.4 and 5). The results suggest that the full WPS benefit is almost achievable despite the presence of hydrogen. In fact, on average toughness values measured in pre-loaded charged specimens were as high as 90% of K_{mat} and 83% of CTOD values from specimens receiving proof loading only.

The analysis given in Table 4-6 considers the use of theoretical models to predict WPS benefit in the fracture toughness of both uncharged and charged specimens at -170°C. The models could not be applied to the test results at -100°C because the pre-load levels were outside the validity limits.

Table 4 compares the measured fracture toughness (K) values against the WPS predictions for the uncharged specimens. This gives an average conservatism of 25% for the Smith's model and 35% for the Chell and Haigh model. For the case of charged specimens, the predictions given in Table 5 assumes that the baseline fracture toughness is the same as measured for uncharged specimens. In this case, the Smith's model gives on average 13% conservative predictions while the Chell and Haigh model is about 22% conservative. In order to make some allowance for the effect of hydrogen on the baseline toughness, the predictions were refined in Table 6 by taking this as 91% of measured uncharged value. This assumption was necessary because there was insufficient material to carry out relevant tests -170°C. The figure of 91% comes from the earlier tests conducted at -100°C. The results given in Table 6 show improved conservatism in the models at an average of 18% for the Smith model and 30% for the Chell and Haigh model.

In general, both models gave on average conservative predictions for both uncharged and charged specimens. However, the predictions for uncharged specimens were more conservative than those for charged specimens. Also, the results suggest that the level of conservatism in the predictions for charged specimens is improved if allowance is made for the effect of hydrogen on the baseline fracture toughness.

It is worth noting that no specific allowance was made for the effect of hydrogen (ageing) in the development of the WPS predictive models. However, it is commonly assumed that the models are applicable provided the baseline toughness adopted takes into account the effect of ageing/embrittlement (see Ref.7). The results of the present study are in line with this assumption.

This study was based entirely on tests conducted on parent metal. However, with regard to the effect of hydrogen on fracture toughness, Davey and Carne[8] found a 40% reduction in upper shelf (CTOD) toughness of specimens made from stove pipe welds. Therefore the effects seen in this study can be expected to be repeated in weld metals.

5. CONCLUSIONS

Fracture toughness tests were conducted on A533B steel specimens following various treatments of hydrogen charging or proof testing or both. The results of the tests have been comprehensively analysed and the following conclusions are drawn:

- Hydrogen charging reduced the baseline fracture toughness of the material. The CTOD was on average reduced by 35% while the reduction in K was smaller at about 9%, for tests carried out at -100°C.

- Significant benefit of warm prestressing was observed in both uncharged and charged specimens at -170°C. The results suggest that most of the WPS benefit (up to 90% of K_{mat} in these tests) is retained after hydrogen charging. However, the proof load levels used in the tests carried out at -100°C were too low to give any significant residual benefit after hydrogen charging.

- The accuracy of WPS predictive models appears to be slightly affected by hydrogen charging. In general, the level of conservatism in the predictions were less for charged specimens than for uncharged ones. However, the predictions for both uncharged and hydrogen-charged specimens were on average conservative.

- The level of conservatism in WPS predictions for hydrogen charged specimens is improved if allowance is made for the effect of hydrogen on the baseline fracture toughness.

- Overall, the results suggest that existing theoretical models by Smith[1,2] and Chell and Haigh[3] may be used to predict fracture toughness even after exposure to hydrogen. However, it should be recognised that the margins on these predictions may be slightly reduced compared to the cases without hydrogen effects. Using reduced baseline toughness values and the appropriate yield properties for the hydrogen-embrittled state should maintain good safety margins in the predictions.

6. REFERENCES

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Table 1 Chemical analyses of A533B class 1 steel (% weight)

C	S	P	Si	Mn	Ni	Cr	Mo	V	Cu	Nb	Ti	Al	B	Sn	Co	CE*
0.18	0.005	0.006	0.24	1.41	0.56	0.18	0.48	<0.002	0.12	<0.002	<0.002	0.018	<0.0003	0.01	0.01	0.59

*CE = C + Mn/6 + (Ni + Cu)/15 + (Cr + Mo + V)/5

Table 2 Summary of fracture toughness results at -100°C

Specimen (M02)		Pre-load condition				Fracture detail		
Set	No.	Pre-load, max (kN)	K ($Nmm^{-3/2}$)	CTOD (mm)	Charging	K_{mat} ($Nmm^{-3/2}$)	CTOD (mm)	Type of result
A	01	0	0	0	None	3934	0.30	c
A	02	0	0	0	None	3525	0.08	c
A	03	0	0	0	None	3809	0.18	c
A	04	0	0	0	None	3893	0.20	c
A	05	0	0	0	None	3594	0.09	c
B	06	0	0	0	Hydrogen	3194	0.06	c
B	07	0	0	0	Hydrogen	3457	0.14	u
B	08	0	0	0	Hydrogen	3467	0.13	m
B	09	0	0	0	Hydrogen	3516	0.11	c
B	10	0	0	0	Hydrogen	3466	0.11	c
C	11	40.5	2884	0.083	[b]Hydrogen	3372	0.08	u
C	12	41.5	2977	0.083	[b]Hydrogen	3367	0.18	u
C	13	44.2	3105	0.087	[b]Hydrogen	3542	0.07	c
C	14	42.5	3080	0.083	[b]Hydrogen	3575	0.11	u
C	15	43.3	3025	0.086	[b]Hydrogen	3512	0.10	u
D	16	40.0	2916	0.079	[a]Hydrogen	3377	0.11	m
D	17	41.3	2934	0.082	[a]Hydrogen	3491	0.09	m
D	18	41.5	2966	0.082	[a]Hydrogen	3532	0.07	c
D	19	42.7	3029	0.084	[a]Hydrogen	3461	0.11	c
D	20	42.0	2949	0.084	[a]Hydrogen	3378	0.05	c
E	21	43.4	3040	0.085	None	3951	0.20	c
E	22	44.5	3052	0.086	None	4091	0.37	u
E	23	44.3	3082	0.085	None	3937	0.14	c
E	24	43.2	3012	0.084	None	3899	0.17	c
E	25	44.3	3065	0.084	None	4014	0.19	c

Notes: [b] = hydrogen charged before proof loading; [a] = hydrogen charged after proof loading
Result type: c = cleavage, u = fracture after tearing, m = maximum load

Table 3 Summary of fracture toughness results at -170°C

Specimen (M03)		Pre-load condition				Fracture detail		
Set	No.	Pre-load, max (kN)	K (Nmm-3/2)	CTOD (mm)	Charging	K_C ($Nmm^{-3/2}$)	CTOD (mm)	Type of result
F	11	0	0	0	None	1322	0.005	c
F	12	0	0	0	None	1540	0.006	c
F	13	0	0	0	None	1957	0.011	c
F	14	0	0	0	None	1746	0.008	c
F	15	0	0	0	None	1110 **Avg = 1535**	0.003 **Avg = 0.007**	c
G	06	44.9	3102	0.093	Hydrogen	2545	0.021	c
G	07	45.1	2991	0.084	Hydrogen	2228	0.012	c
G	08	45.1	2983	0.080	Hydrogen	2068	0.013	c
G	09	45.0	2965	0.077	Hydrogen	2108	0.014	c
G	10	45.0	3015	0.086	Hydrogen	2866 **Avg = 2363**	0.027 **Avg = 0.017**	c
H	01	44.5	3041	0.086	None	3182	0.029	c
H	02	44.9	2966	0.083	None	3263	0.030	c
H	03	44.2	3054	0.097	None	2516	0.019	c
H	04	44.3	3035	0.087	None	2142	0.013	c
H	05	45.1	3017	0.090	None	1968 **Avg = 2614**	0.011 **Avg = 0.020**	c

Notes: Result type: c = cleavage, u = fracture after tearing, m = maximum load
Avg. = average value

Table 4 Comparison of measured toughness against theoretical predictions for uncharged specimens

Specimen (M03)		Measured toughness, K_{mat}	Preload K= K_1 (kN)	*Assumed base toughness, K_C ($Nmm^{-3/2}$)	Proof load ratio K_1/K_C	Smith Model Predictions		Chell and Haigh Model Predictions	
Set	No.					Failure K_f	Act./pred. (K_{mat} /K_f)	Failure K_f	Act./pred. (K_{mat} /K_f)
H	01	3182	3041	1535	1.98	2094	1.52	1944	1.64
H	02	3263	2966	1535	1.93	2066	1.58	1929	1.69
H	03	2516	3054	1535	1.99	2099	1.20	1946	1.29
H	04	2142	3035	1535	1.98	2092	1.02	1943	1.10
H	05	1968	3017	1535	1.97	2085	0.94 **Avg = 1.25**	1939	1.02 **Avg = 1.35**

Note: * K_c taken as average of uncharged non-proof loaded toughness values (i.e., set F average)

Table 5 Measured toughness versus theoretical prediction for charged specimens (Baseline toughness takes no account of hydrogen)

Specimen (M03)		Measured toughness, K_{mat}	Preload K= K_1 (kN)	*Assumed base toughness, K_C ($Nmm^{-3/2}$)	Proof load ratio K_1/K_C	Smith Model Predictions		Chell and Haigh Model Predictions	
Set	No.					Failure K_f	Act./pred. (K_{mat} /K_f)	Failure K_f	Act./pred. (K_{mat} /K_f)
G	06	2545	3102	1535	2.02	2116	1.20	1956	1.30
G	07	2228	2991	1535	1.95	2075	1.07	1934	1.15
G	08	2068	2983	1535	1.94	2072	1.00	1932	1.07
G	09	2108	2965	1535	1.93	2066	1.02	1929	1.09
G	10	2866	3015	1535	1.96	2084	1.38 **Avg = 1.13**	1939	1.48 **Avg = 1.22**

Note: * K_c taken as average of uncharged non-proof loaded toughness values (i.e., set F average)

Table 6 Measured toughness versus theoretical prediction for charged specimens (baseline toughness allows for hydrogen effect)

Specimen (M02)		Measured toughness, K_{mat}	Preload K= K_1 (kN)	*Assumed base toughness, K_C ($Nmm^{-3/2}$)	Proof load ratio K_1/K_C	Smith Model Predictions		Chell and Haigh Model Predictions	
Set	No.					Failure K_f	Act./pred. (K_{mat} /K_f)	Failure K_f	Act./pred. (K_{mat} /K_f)
G	06	2545	3102	1397	2.22	2030	1.25	1836	1.39
G	07	2228	2991	1397	2.14	1988	1.12	1814	1.23
G	08	2068	2983	1397	2.14	1985	1.04	1812	1.14
G	09	2108	2965	1397	2.12	1979	1.07	1808	1.17
G	10	2866	3015	1397	2.16	1997	1.44 **Avg = 1.18**	1818	1.58 **Avg = 1.30**

Note: * K_c taken as 91% of average uncharged baseline toughness value

TWI ENDORSEMENT

This work has been carried out in accordance with TWI's QA Procedures.

Project Leader ....

Head of Department ....
(or delegate)

Figure 1: Load-clip trace for a typical preload

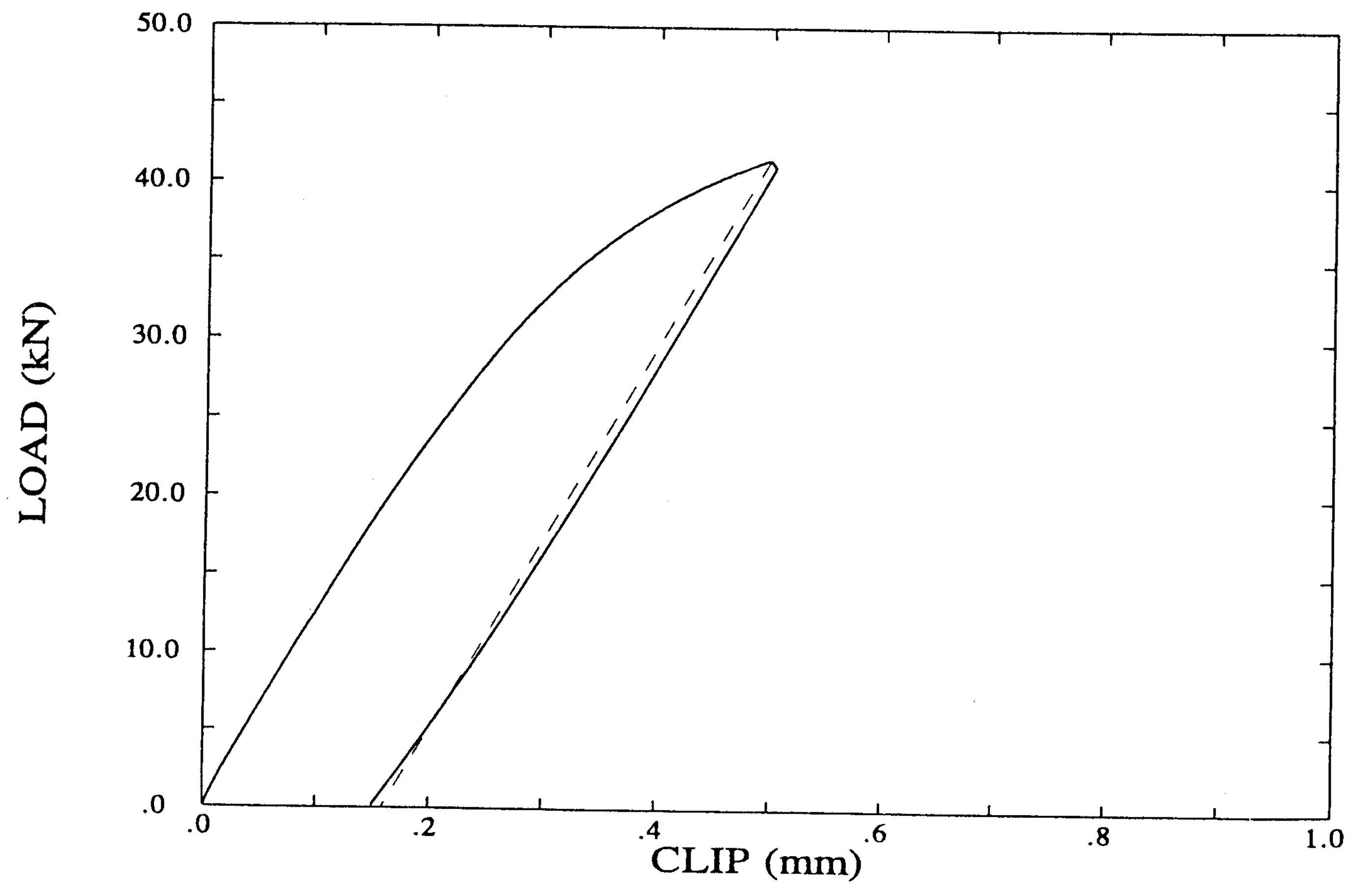

Figure 2: Fracture toughness (K) values from tests at -100 deg.C

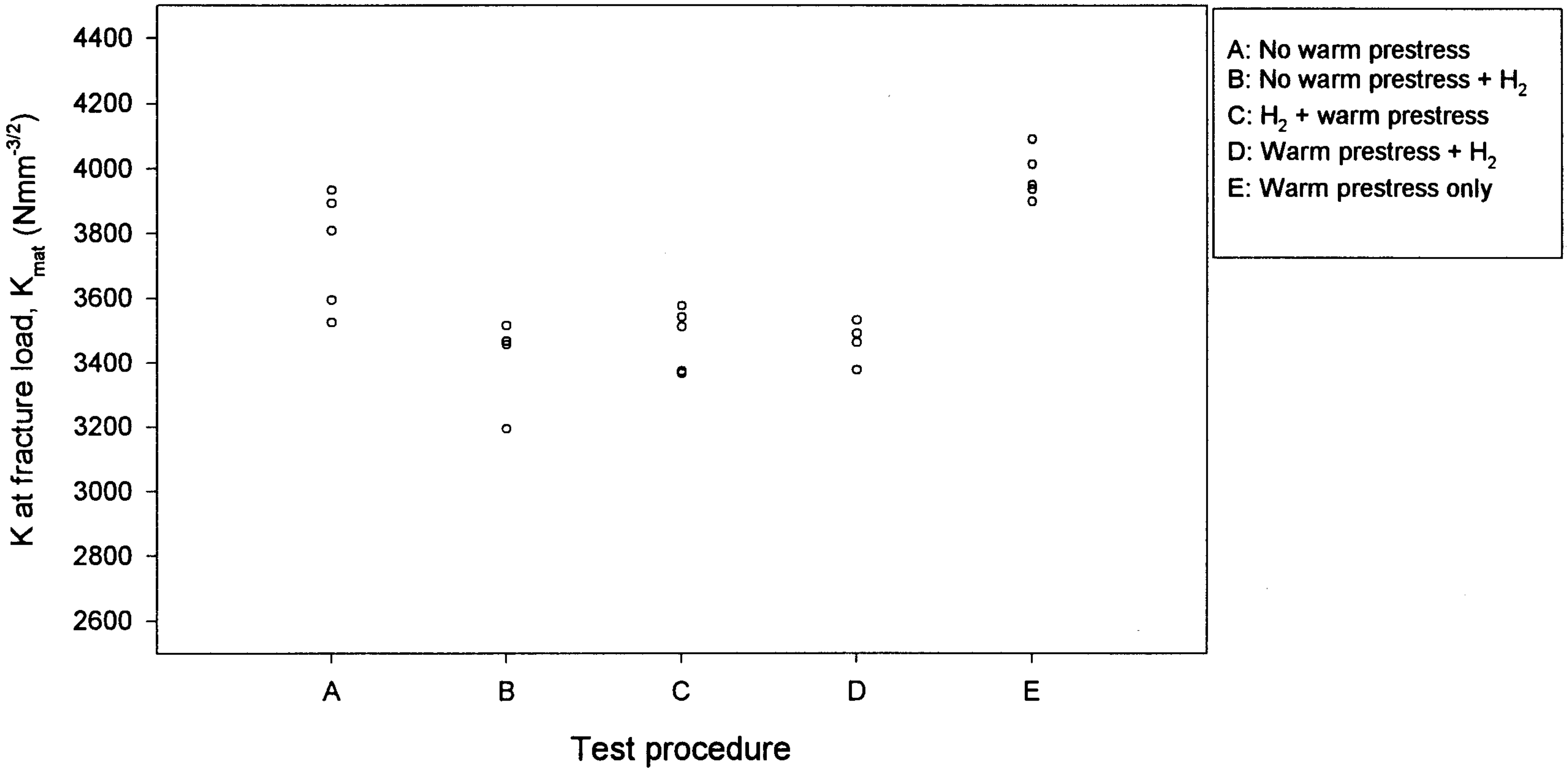

Figure 3: CTOD from fracture toughness tests at -100 deg.C

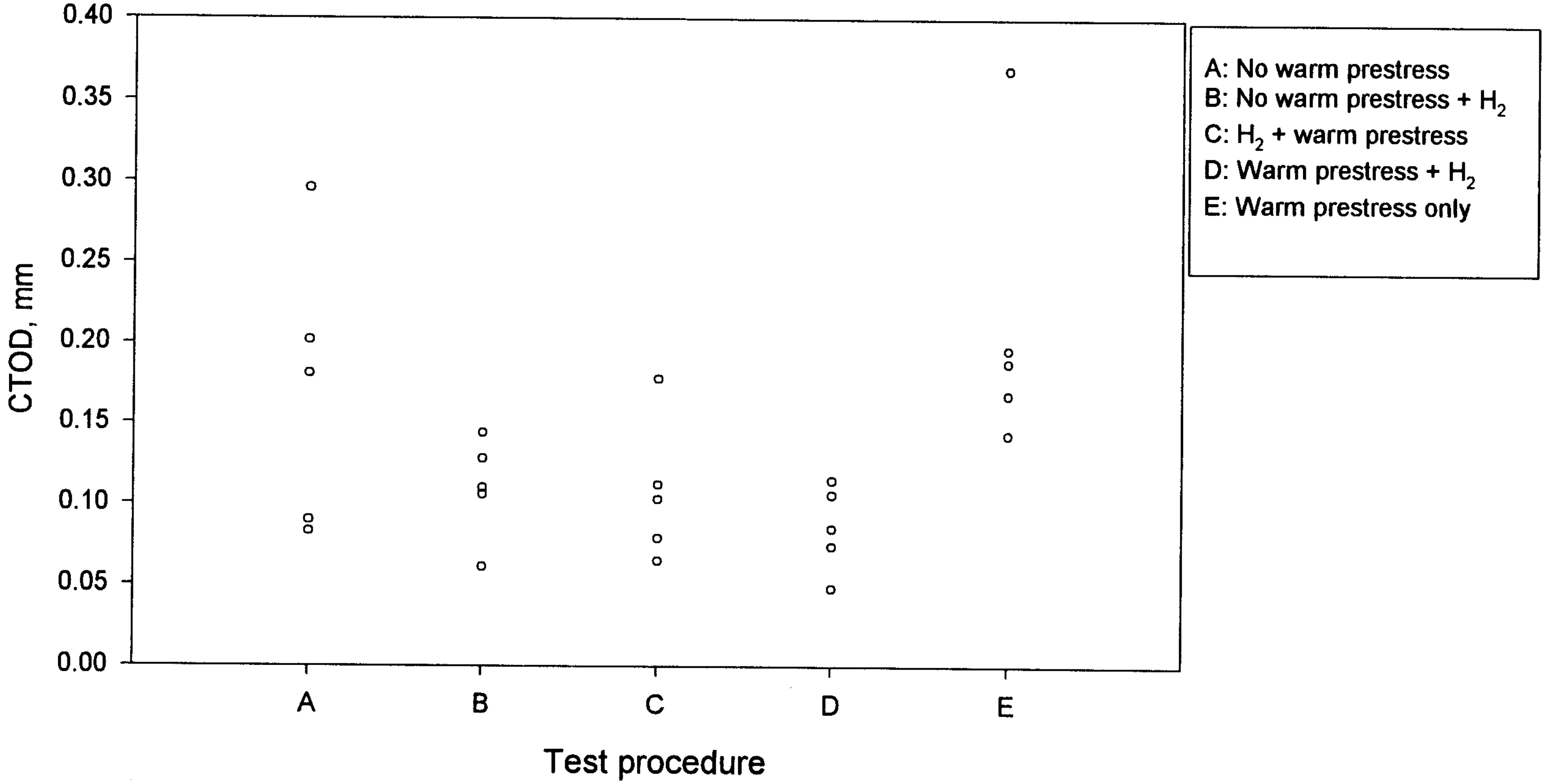

Figure 4: Fracture toughness (K) values from tests at -170 deg.C

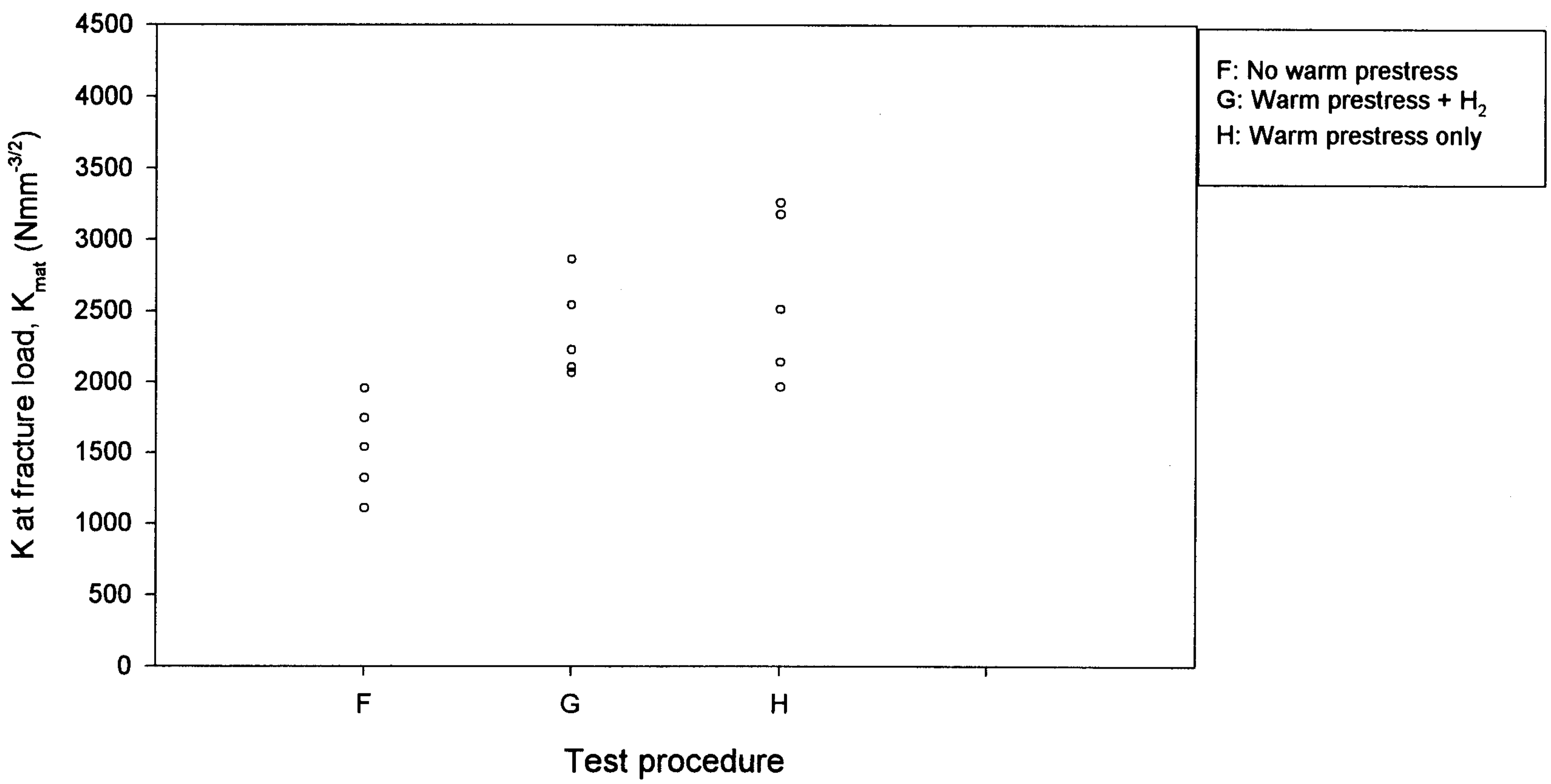

Figure 5: CTOD values from fracture toughness tests at -170 deg.C