



EUROPEAN
COMMISSION

Community Research



MATISSE

Materials' Innovations for Safe and Sustainable nuclear in Europe

**Support to the development of joint research actions
between national programmes on advanced nuclear materials**

FP7-Fission-2013

Combination of Collaborative project (CP) and Coordination and Support Actions (CSA)

Grant agreement no: 604862

Start date: 01/11/2013 Duration: 48 Months

D.5.13

***A synthesis report on developed and applied
computational models for creep-fatigue crack
growth with comparison of predictions with tests
performed and recommendations for engineering
applications***

MatISSE – Contract Number: 604862

Document title	A synthesis report on developed and applied computational models for creep-fatigue crack growth with comparison of predictions with tests performed and recommendations for engineering applications
Author(s)	Karl-Fredrik Nilsson (JRC), Matthias Bruchhausen (JRC), Mara Serrano (CIEMAT), Antonio Muñoz (CIEMAT), Rebeca Hernandez (CIEMAT), Hyeong-Yeon Lee (KAERI), Woo-Gon Kim (KAERI), Jaromir Janoušek (CVR)
Number of pages	70
Document type	Deliverable
Work Package	5
Document number	D5.13
Issued by	JRC
Date of completion	12/02/2018
Dissemination level	Public

Summary

The report summarized the three different studies related to high temperature structural integrity: creep-fatigue crack propagation tests and analysis following ASTM E2760 with the objective to evaluate the Standard and derive creep-fatigue crack propagation data for P91 and 316L; thermal-fatigue tests of 316L and P91 pipe components as a benchmark for typical service-exposed load conditions; creep crack propagation, tensile and fracture toughness tests of P91 welds and service exposed material.

Approval

Rev.	Date	First author	WP leader	Project Coordinator
0	02/2018	K.F. Nilsson, JRC	K.F. Nilsson, JRC	P.F. Giroux, (CEA)
		05/02/2018	05/02/2018	12/02/2018
				

Distribution list

Name	Organisation	Comments
All beneficiaries	MatISSE	

Table of contents

1	Introduction	5
2	Background Creep-Fatigue assessment	6
3	Creep-Fatigue Crack Growth Testing	11
3.1	ASTM Test Procedure	11
3.1.1	Specimen.....	11
3.1.2	Test Machine and Loading	11
3.1.3	Heating	11
3.1.4	Measurements.....	12
3.1.5	Functional relationships and calculations.....	12
3.1.6	Presentation of results.....	13
3.2	JRC Creep fatigue tests.....	13
3.2.1	JRC Test Procedure and Test Matrix	13
3.2.2	Results.....	15
3.2.3	Concluding remarks JRC tests.....	23
3.3	CIEMAT Creep-fatigue tests.....	24
3.3.1	Introduction.....	24
3.3.2	Materials	24
3.3.3	Test specimen	24
3.3.4	Experiments and test procedure	25
3.3.5	Analysis Procedure.....	26
3.3.6	Results.....	27
3.4	CVR creep-fatigue tests.....	40
4	JRC Thermal Fatigue Tests	44
4.1	Introduction	44
4.2	Experimental Programme	44
4.3	Modelling	46
4.3.1	Thermal analysis	46
4.3.2	Stress Analysis	47
4.3.3	Fracture Analysis.....	48
4.4	Discussion	51
5	KAERI Tests Creep crack Growth, Tensile and Fracture Toughness P91 (base, weld and aged)	52
5.1	CCG test results of welded Gr.91 specimens at 550°C and 600°C	52
5.1.1	Basic material tests associated with CCG tests.....	52
5.1.2	Determination of material constants at 550°C and 600°C	54

5.1.3	Calculation of C^* fracture parameter	56
5.1.4	Construction of CCG laws at 550 °C	57
5.1.5	Construction of CCG laws at 600° C	58
5.2	High temperature material strength tests for Gr.91 steel	59
5.2.1	Virgin Gr.91 Specimens	59
5.2.2	Service-exposed Gr.91 steel	60
5.2.3	Material strength tests	60
5.2.4	Comparison of tension test results for virgin and service-exposed Gr.91 steel	63
5.3	J-R tests for Gr.91 steel	64
5.3.1	J-R test details	65
5.3.2	Comparison of J-R test results for virgin and service-exposed Gr.91 steel specimens	65
6	Concluding Remarks	68
7	References	70

1 Introduction

In liquid metal fast reactors the pressure is low and the dominant loads are thermo-mechanical from planned operation actions such as start-ups and shutdowns and in the future also load following. There are also thermo-mechanical loads that results from the normal operational conditions but difficult to control such as i) stratification of fluid with a hot and cold layer or ii) thermal striping at mixing tees or a moving liquid surface. All these load cases induce cyclic loads and where the thermal loads induce thermal stresses and strains with peak values at the surface of components and stress amplitudes then decay away from the surface. Since the maximum amplitude of stresses and strains can be quite high, surface cracking may appear after relatively short time. Such surface cracking will only marginally affect the structural integrity of a component. The safe life-time of the component is therefore controlled by the propagation from a surface crack into a through-wall crack. Thus the understanding and prediction of cracks under creep-fatigue conditions is crucial for the assessment of nuclear components.

The report consists of the following parts:

- The creep-fatigue crack propagation is much more complex than pure fatigue and creep crack growth and a short summary of the basic concepts and specific complexities related to the creep-fatigue interaction are outlined.
- Due to the complexities of creep-fatigue crack propagation the test standard ASTM E2760-16 [1], which was first published in 2010. The main part of the work has been to perform creep-fatigue tests using Compact Tension fracture specimens and their assessment based on this standard by JRC, CIEMAT and CVR. The main features of the standard E2760-16 are outlined and the results by each partner are reported separately. All three partners tested Gr. 91 but CIEMAT also performed tests on 316L.
- KAERI made investigated the creep crack growth rates (CCGR) for base metal (BM), weld metal (WM), and heat affected zone (HAZ) at 550°C and 600°C in the weldment of Gr.91, which is prepared using a shielded metal arc weld (SMAW) process. Creep crack growth laws at 550°C and 600°C for the BM, WM, and HAZ are newly constructed using a fracture parameter of C^* . The tested CCGR lines were then compared with RCC-MRx code. KAERI also tensile tests and fracture toughness tests of virgin and service-exposed Gr 91 steel.
- The fatigue life assessment for components subjected to thermo-mechanical loads is more complicated and a combined experimental and numerical assessment of a pipe components of both P91 and 316L subjected to thermal fatigue is summarized.
- Finally the different test results and analyses are briefly discussed and compared. Some conclusions are drawn and recommendations for future work are given.

2 Background Creep-Fatigue assessment

The crack propagation in pure fatigue is transgranular by alternating slip whereas the mechanism for creep crack propagation is intergranular by void nucleation and coalescence. In creep-fatigue there is a combination of both. Fatigue crack propagation in absence of hold times as well as pure creep crack propagation under steady-state conditions are rather well understood and can be predicted with good accuracy using fracture mechanics. Much of the work in high-temperature creep and fracture were established by Riedel [2] and in particular the creep-fatigue crack propagation is much due to Saxena and co-workers [3-7]. The starting point in creep-fatigue crack propagation is that crack propagation can be separated into a time-independent fatigue part and a time dependent creep part:

$$\frac{da}{dN} = \left(\frac{da}{dN}\right)_f + \left(\frac{da}{dN}\right)_c. \quad (2.1a)$$

Creep-fatigue crack propagation involves both trans- and intergranular damage whereas the separation assumes that the creep and fatigue parts are uncoupled. Another approach proposed by Saxena is that trans- and intergranular crack growth are competing mechanism and creep-fatigue growth is governed by the dominating damage mechanism [6,7], i.e. $\frac{da}{dN} = \max\left[\left(\frac{da}{dN}\right)_f, \left(\frac{da}{dN}\right)_c\right]$.

The fatigue part is governed by Paris law,

$$\left(\frac{da}{dN}\right)_f = B \cdot \Delta K^m, \quad (2.1b)$$

and the creep part is computed by integrating the crack propagation rate, which is controlled by an energy flux integral $C(t)$, over the hold time.

$$\left(\frac{da}{dN}\right)_c = \int_0^T \frac{da}{dt} dt \left\{ \begin{array}{l} \frac{da}{dt} = Q \cdot C(t)^q \end{array} \right\}. \quad (2.1c)$$

Here B , m and Q , q are the fatigue and creep crack propagation material properties derived from standard tests, ΔK is the range of the stress intensity factor and $C(t)$ is the energy flux integral which depends on the visco-elastic properties, the crack geometry and the time. Mathematically the $C(t)$ integral is defined by:

$$C(t) = \int_{\Gamma \rightarrow 0} W^* dy - T_i \left(\frac{\partial \dot{u}_i}{\partial x} \right) ds. \quad (2.2)$$

W^* is the stress power and T_i the traction vector and $\frac{\partial \dot{u}_i}{\partial x}$ the strain rate along a contour very close to the crack tip. Computation of the $C(t)$ -integral requires a proper constitutive model, which in the creep-fatigue cases means a cyclic visco-plastic model. The creep material properties have a strong temperature dependency. From the definition in Eq. 2.2 it follows that $C(t)$ is controlled by the stresses and strains at the crack tip and the integral contour should only enclose the area at the crack tip. A steady state creep situation may prevail after long times and the C -integral reaches then an asymptotic value $\lim_{t \rightarrow \infty} C(t) = C^*$, which is independent of the crack contour and can therefore be computed relatively easily by finite elements or derived from formulas. The value of $C(t)$ depends strongly on the creep properties.

The situation becomes much more complicated in creep-fatigue, and a major reason is that the hold times are generally much shorter than the time needed to reach steady-state creep, and we are therefore faced with a

situation of cyclic creep transients. Other complicating factors that need to be addressed are the visco-plastic properties e.g. cyclic hardening or softening that depend on the local stresses and the incubation time for onset of creep-crack growth. Moreover cyclic softening also strongly increases the creep rate.

If a cracked body is exposed to a step load then for a stationary crack the computed energy flux decays monotonically with time as the stresses at the crack tip relax and the creep zone emanating from the crack tip grows as illustrated schematically in **Figure 2.1**. In the case that the creep curve has a well-defined secondary creep regime, the C-integral reaches a steady state value, $\lim_{t \rightarrow \infty} C(t) = C^*$. At the initial stage, small-scale creep, the creep zone is much smaller than the defect and the plastic zone at the crack tip, and assuming secondary creep ($\dot{\epsilon} = A\sigma^n$), the C-integral can be estimated from the stress intensity factor derived from an elastic analysis,

$$C_{ssc}(t) = \frac{K^2(1-\nu^2)}{(n+1)Et}. \quad (2.3)$$

In-between the small scale creep and the steady state creep we have a transient creep. For long times the $C(t)$ must approach the steady-state value C^* . By approximating the $C(t)$ as the sum of the small scale creep and steady state solution an analytical expression can be obtained for the C-integral.

$$C(t) \approx C_{ssc}(t) + C^* = \left(1 + \frac{t_T}{t}\right) \cdot C^* \quad (2.4)$$

For the creep-fatigue crack propagation assessment it is important to classify the creep regime and a "transition time", t_T , which is defined from the conditions that $C_{ssc}(t) = C^*$. It then follows directly from Equation (2.3) and (2.4) that the transition time is:

$$t_T = \frac{K^2(1-\nu^2)}{(n+1)EC^*}, \quad (2.5)$$

where E and ν are Young's modulus and Possion's ratio respectively and n the exponential term in a Norton secondary creep law. If the hold time, t_h , in a creep-fatigue test is much longer than the transition time, t_T , then the C^* can be used. The stress intensity factor can be directly calculated from standard formulas for specific geometries. An expression for the C^* integral for the CT specimen under secondary creep is given by [1]:

$$C^* = A(W-a)h_1\left(\frac{a}{W}, n\right)\left(\frac{P}{1.455\eta_1 B(W-a)}\right)^{n+1} \left\{ \eta_1 = \left[\left(\frac{2a}{W-a}\right)^2 + 2\left(\frac{2a}{W-a}\right) + 2 \right]^{1/2} - \left[\left(\frac{2a}{W-a}\right) + 1 \right] \right\} \quad (2.6)$$

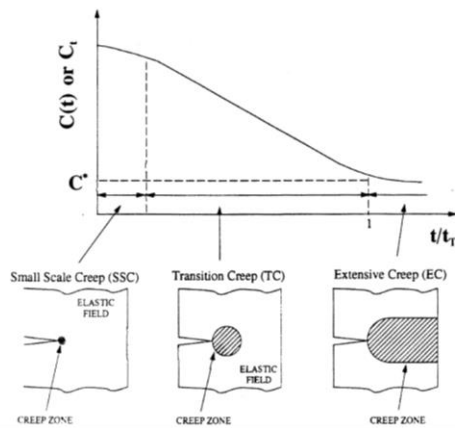


Figure 2.1 Typical evolution of the $C(t)$ integral for a stationary crack during small-scale, transition and extensive creep [2].

For a ferritic-martensitic steel, such as P91, the creep rate versus total creep is U-shaped: a primary creep for which the creep rate decreases, and a tertiary creep for which the creep rate increases so there is a minimum creep rate, but no secondary creep regime as illustrated in **Figure 2.2a**. The minimum creep rate occurs typically at 1% creep strain. Using the minimum creep rate for A and n would drastically overestimate the transition time and underestimate the C^* value. Moreover there is also significant cyclic softening (**Figure 2.2b**) which also may increase the creep rate dramatically (**Figure 2.2c**). This complex visco-plastic behaviour renders ferritic-martensitic steels particularly difficult to analyse.

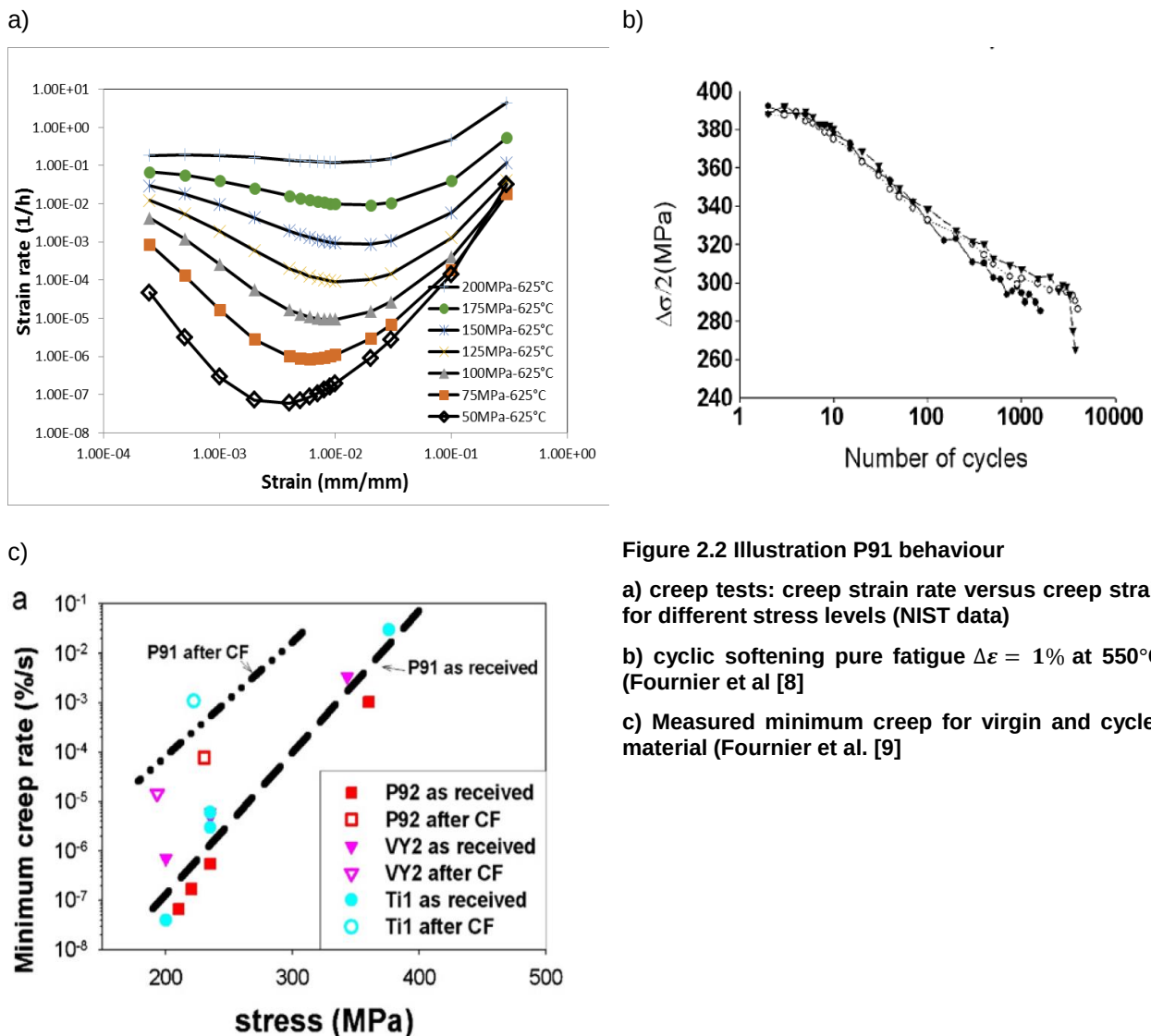


Figure 2.2 Illustration P91 behaviour

a) creep tests: creep strain rate versus creep strain for different stress levels (NIST data)

b) cyclic softening pure fatigue $\Delta\varepsilon = 1\%$ at 550°C . (Fournier et al [8])

c) Measured minimum creep for virgin and cycled material (Fournier et al. [9])

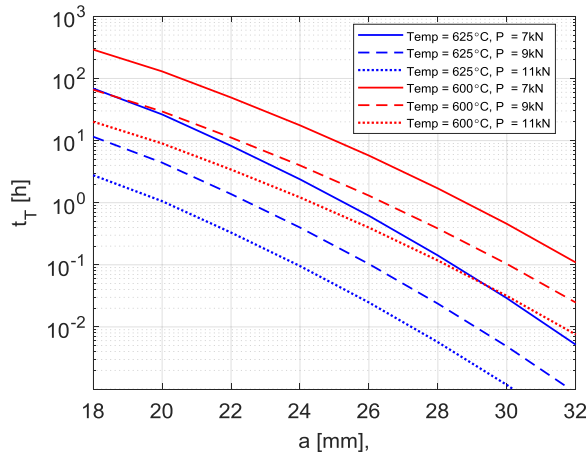


Figure 2.3 Computed transition time to steady-state creep crack growth for CT-specimen and P91 using average creep rate up to rupture from pure creep tests.

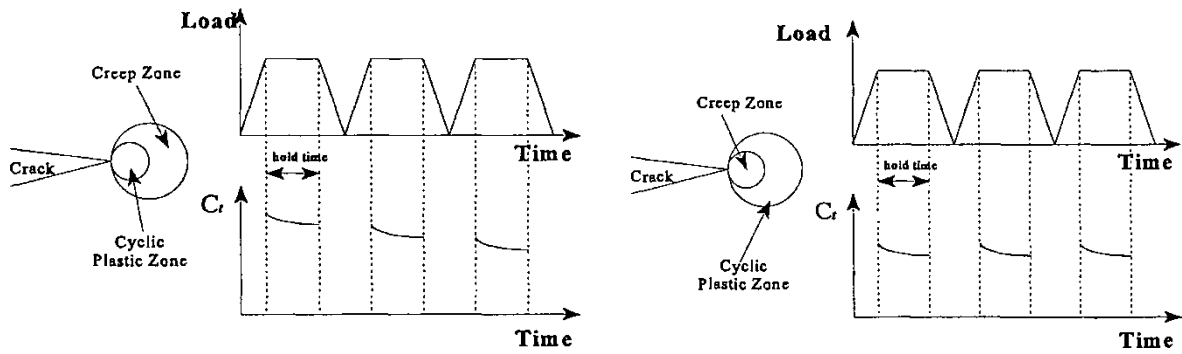


Figure 2.4 Schematic illustration of the $C(t)$ integral versus time for load controlled test a) creep zone larger than cyclic plastic zone b) creep zone smaller than cyclic plastic zone at crack tip [5].

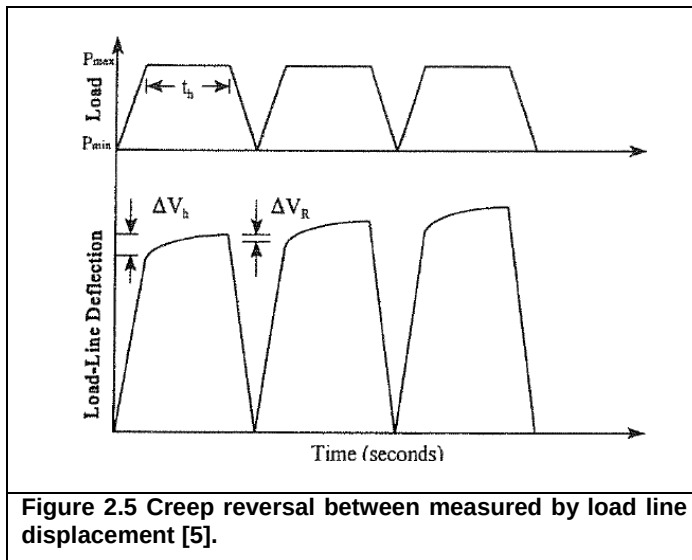


Figure 2.5 Creep reversal between measured by load line displacement [5].

Figure 2.3 shows the computed transition time for versus the crack size for the Standard CT-specimen for P91 data at 600°C and 625°C, and specimen size $W = 50$ mm and $B = 12.5$ mm, for the three load levels 7, 9 and 11 kN, and where the C^* for the CT-specimen is given by Eq. 2.6. In the analysis A and n are calibrated from creep rupture tests but where A and n have been fitted to rupture data, which should be a conservative estimate for pure creep. The creep rate can also be non-conservative if the material has undergone cyclic softening as seen in **Figure 2.2c**. The transition times are strongly dependent on the temperature, defect size and applied load and may differ by order of magnitude between test conditions in a test programme.

Close to the crack tip stresses and strains will include high instantaneous plastic and time dependent creep deformation. The interaction between creep and plastic deformation, and in particular the size of the creep and cyclic plastic zones, control the creep-fatigue crack growth. It was mentioned above that the C -integral has its largest value when small scale creep conditions prevail. Plastic deformation reduces the stress at the crack tip and hence reduces the creep rate. Thus there is an intricate interaction between the creep and plastic deformation at the crack tip. If the creep zone is much larger than the plastic zone at the crack tip, then it is expected that the crack growth in creep fatigue develops monotonically between cycles whereas if the creep zone is much smaller than the plastic zone, the creep at the crack tip may be reversed and the small-scale creep rate is re-initiated for

each cycle, as illustrated in **Figure 2.4**. During cyclic loading some of the accumulated creep at the crack tip will also be reversed. From a creep-fatigue crack propagation tests this is quantified by the difference in the measured load line displacement at the end and beginning of the hold time for two consecutive cycles, ΔV_R [5] as illustrated in **Figure 2.5**. The creep reversal factor, $C_R = \Delta V_R / \Delta V_h$, where ΔV_h is the load line displacement

during the hold time. A small creep reversal factor indicates large scale creep and a large value is associated with small-scale creep.

When a component or specimen with a sharp crack is subjected to a step load, the crack initially blunts and onset of creep crack propagation will only take place after sufficient creep damage has occurred at the crack tip and the crack tip has reached a critical crack opening displacement [10]. For a creep-fatigue test it would have less importance but may need to be taken into account for at least the first load cycles.

3 Creep-Fatigue Crack Growth Testing

3.1 ASTM Test Procedure

The creep-fatigue test Standard ASTM E2780 "Standard Test for Creep-Fatigue Crack Growth Testing" was first issued in 2010 and with a revised version in 2016 [1]. CVR, JRC and CIEMAT followed this Standard for their tests and a brief summary is given below.

3.1.1 Specimen

Only the CT-specimen is recommended with dimensions shown in **Figure 3.1**. The recommended width-to-thickness ratio (W/B) is 4 and should not exceed 8. The initial crack size, a_0 , including starter notch and pre-cracking, should be between $0.25W$ and $0.35W$.

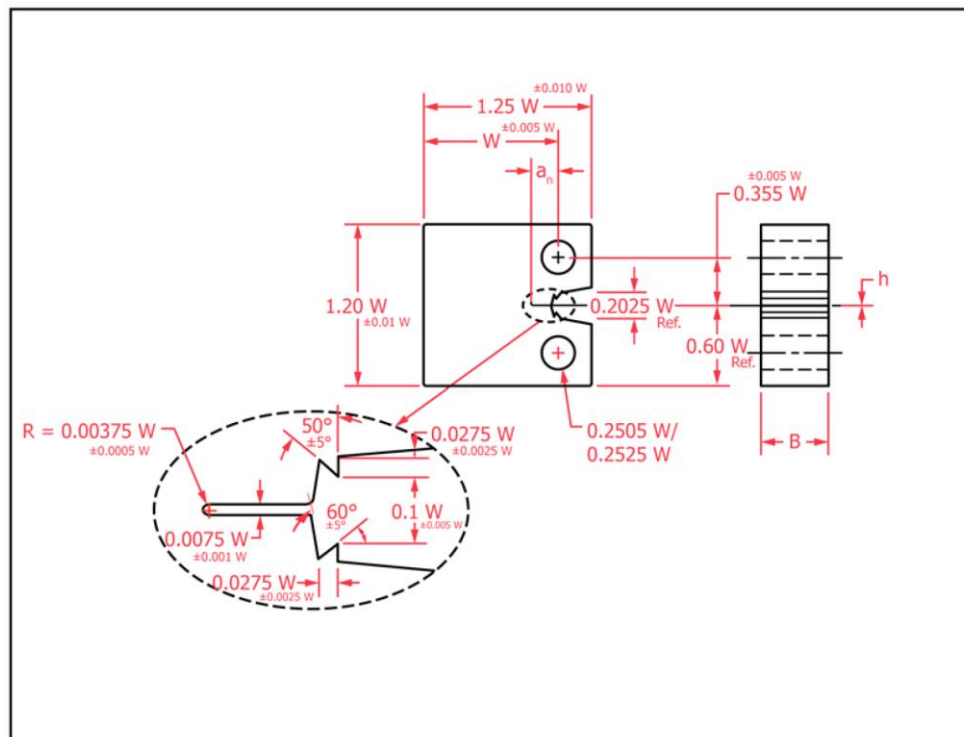


Figure 3.1 Recommended Specimen CT-specimen geometry (ASTM E2780-16)

3.1.2 Test Machine and Loading

Test shall be conducted using servo-controlled tension-compression fatigue machine, but hydraulic and electromechanical machines are also acceptable. The loading is done in force-control and forces need to be measured by an accuracy better than 1%. Good alignment of the force-line is essential. The load cycles can be: a) low frequency triangular wave forms b) saw-tooth wave forms or c) cyclic hold forms comprising a series of ramps and hold-time at the maximum load. The latter is the wave form used by the MASISSE partners.

3.1.3 Heating

Heating shall be by electric resistance or radiation furnace. The deviation between indicated temperature and nominal test temperature should not exceed $\pm 2^\circ\text{C}$.

3.1.4 Measurements

Continuous force-line displacement measurement as function of creep-fatigue cycles is necessary for the evaluation. Knife edges are recommended for friction free-seating of the gage. The accuracy of the displacement gage should be within $\pm 2\%$.

A direct current (DCPD) or alternating current (ACDP) electric potential drop crack monitoring system should be used. The PD system should be capable of resolving crack extension of at least $\pm 0.1\text{mm}$.

Temperature measurement and control should be done using Class 1 thermocouples.

An automated digital recording system should be used to simultaneously collect and process the force, the force-line displacement, the PD signal and the temperature as function of time and cycles. The sampling frequency should be sufficient to correctly define the loading cycles. In particular start and end of cycles and hold times must be included. It is recommended to have 200 points for the loading and unloading portions of the cycle and 100-200 point during each hold time.

3.1.5 Functional relationships and calculations

The crack length, a , is measured from the PD signal and the initial, a_0 , and final, a_f , crack length:

$$a = \left[(a_f - a_0) \frac{(U - U_0)}{(U - U_f)} \right] + a_0, \quad (3.1)$$

where U is the instantaneous potential difference corresponding to a , and U_0 , U_f the initial and final potential difference. The stress intensity factor K is determined from:

$$K = \frac{P}{\sqrt{BB_N W}} f\left(\frac{a}{w}\right) \quad \left\{ f = \left[\frac{2 + \frac{a}{w}}{\left(1 - \frac{a}{w}\right)^{\frac{3}{2}}} \right] \left(0.886 + 4.46 \left(\frac{a}{w}\right) - 13.32 \left(\frac{a}{w}\right)^2 + 14.72 \left(\frac{a}{w}\right)^3 - 5.6 \left(\frac{a}{w}\right)^4 \right) \right\} \quad (3.2)$$

The $C(t)$ integral is a mathematical expression which cannot be directly measured in a test. Instead Saxena introduced the C_t parameter which is determined as the difference in stress power per unit with infinitesimally differing crack areas, which can be measured from a creep fatigue test:

$$C_t = \frac{P \dot{V}_c}{\sqrt{BB_N W}} \left(\frac{f'}{f} \right), \text{ where } f' = \frac{df}{d(a/W)} \quad (3.3)$$

Note that C_t and $C(t)$ are similar but not identical. Under extended creep they are identical and approach C^* , ($\lim_{t \rightarrow \infty} C(t) = \lim_{t \rightarrow \infty} C_t = C^*$). In the assessment of creep-fatigue crack propagation tests a mean value of the load displacement rate and crack length during the hold time is adopted, $(C_t)_{avg} = \frac{1}{t_h} \int_0^{t_h} C_t dt$, which from Eq. (3.3) gives:

$$(C_t)_{avg} = \frac{P \Delta V_h}{\sqrt{BB_N W} t_h} \left(\frac{f'}{f} \right) \quad (3.4)$$

For short hold times the increase in the load-line displacement during hold is often only a very small fraction of the total cyclic load-line displacement, and may be too small to measure accurately. In that case the Standard allows to use an analytical expression:

$$(C_t)_{avg} = \frac{2\alpha\beta(1-\nu^2)}{E} \cdot F_{cr}(\theta, n) \cdot \frac{K_h^4}{E^2 W} \cdot \left(\frac{f'}{f}\right) \cdot \frac{2}{n-1} \cdot (E^n A)^{\frac{2}{n-1}} \cdot t_h^{\frac{n-3}{n-1}} + C^*(t) \quad (3.5)$$

$$\text{where } \alpha = \frac{1}{2\pi} \left[\frac{(n+1)^2}{1.38n} \right]^{\frac{2}{n-1}}.$$

This expression is only valid for secondary creep and previous creep damage is not included (i.e. full load reversal) and should therefore be used with caution. The stress intensity factor and the C^* are given by Eq. (3.2) and (2.6) respectively. A value of $\beta = 0.33$ can be used and the function F_{cr} is tabulated in the Standard.

3.1.6 Presentation of results

The two main ways to present the results are:

- crack growth per cycle (da/dN) versus the stress intensity range (ΔK);
- average time rate of crack growth during hold time $(da/dt)_{avg}$ as function of $(C_t)_{avg}$.

3.2 JRC Creep fatigue tests

3.2.1 JRC Test Procedure and Test Matrix



Figure 3.2.1 CT specimen with side grooves

The material used at JRC was a 60 mm thick plate (heat nr. 20057) from ArcelorMittal. The plate was tested in "as received" product state and was submitted to the following heat treatment: 1060°C – 4Hrs + Water Quenching + 760°C – 3hrs and 20min – Air Cooling. This plate was produced in full accordance with RCC-MRx requirements (STR RM 2432).

All tests use the CT-specimen following the ASTM E2780 requirements. All specimens have the same nominal dimensions: width $w = 50$ mm, thickness $B = 12.5$ mm. A notch with a maximum width of 0.15 mm was machined by electrical-discharge machining (EDM). Fatigue pre-cracking was then used to introduce a sharp starter crack. Notches were introduced after the pre-cracking to achieve a width (B_N) of 10 mm. **Figure 3.2.1** shows a specimen with side grooves.

All tests were conducted using an INSTRON servo-hydraulic test machine and the test specimen heated in a furnace. The specimen was heated to the specified temperature and maintained at this temperature for at least 30 minutes before loading. The displacement was measured by extensometers attached at the knife edges of the specimen (**Figure 3.2.2**). All tests were done in load control under cyclic loading with 2 seconds loading, constant load hold (t_{Hold}), and 2 seconds unloading.

Direct current displacement drop (DCDP) was used to monitor crack growth. To reduce the effect of variations in current and temperature the DCDP signal was measured both on the specimen (X) and a reference sample (Y). The ratio X/Y was used for the crack size measurement. All results were sampled with a frequency of 1 Hz, which allowed a very high resolution of the measured data. The initial, a_0 , and final, a_f , crack lengths were measured by breaking up the specimen at a low temperature under cyclic loading as shown in **Figure 3.2.3**. The crack length is then derived from linear relationship between the crack length and potential drop (Eq 3.1). The final crack front length is measured from the average of five measurement across the specimen thickness.

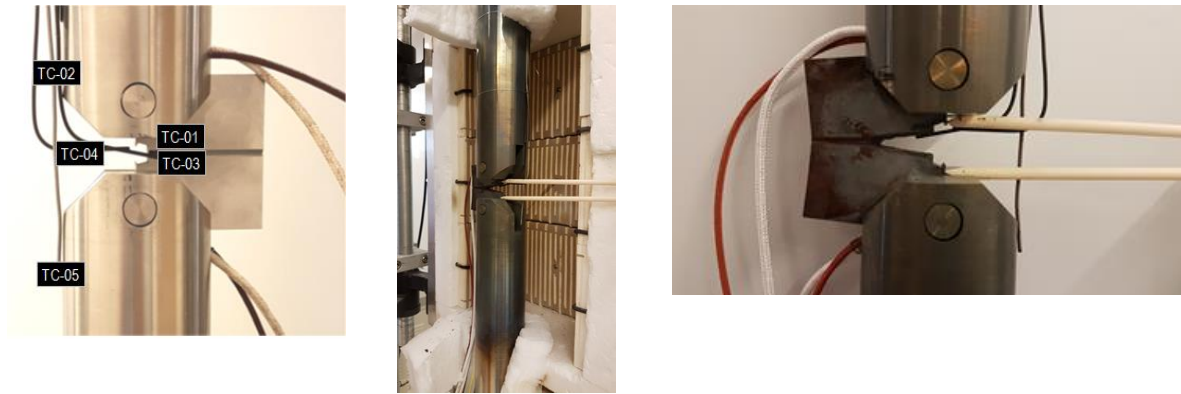


Figure 3.2.2 Specimen with thermocouples; specimen with extensometers

The JRC test matrix is shown in **Table 3.2.1**. The six tests AN-03 to AN-08 used a fixed hold time for each test. The five tests BV-01 to BV-05 all employed a two-block loading consisting of 10 cycles with hold-time 3600 seconds (total duration 36040 seconds or 10h and 40 seconds) and 490 cycles with hold-time 60 seconds (total duration 31360 seconds or 8 h and 2560 seconds). This block loading was repeated until the end of the test.

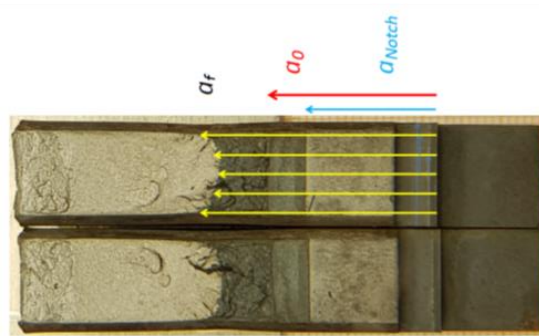


Figure 3.2.3 Specimen broken after test to measure initial and final crack length (BV01).

Table 3.2.1 Summary of JRC tests

Specimen	Temp [°C]	t_{hold} [s]	P(kN)/R	B [mm]	B_N [mm]	W [mm]	a_0 [mm]	a_f [mm]	NF short/long	N_f tot	Time Short/long (h)	Time tot h
AN-03	625	0	9/0.1	12.5	10	50	19.36	31.77	-	13958	-	15.5
AN-04	625	600	9/0.1	12.5	9.75	50	20.01	24.66	-	2211	-	371.0
AN-05	625	60	9/0.1	12.5	10	50	19.58	25.6	-	7105	-	126.3
AN-06	600	0	9/0.1	12.5	10	50	19.59	32.99	-	15827	-	17.6
AN-07	625	60	9/0.1	12.5	10	50	20.35	30.20	-	6777	-	120.5
AN-08	600	60	9/0.1	12.5	10	50	20.35	31.94	-	12525	-	222.7
BV-01	625	10*3600/490*60	9/0.1	12.5	10	50	20.50	27.44	5390/111.17	5501.17	95.8/113.3	207.1
BV-02	625	10*3600/490*60	11/0.1	12.5	10	50	19.67	24.51	1470/34.99	1504.99	26.13/35.0	61.16
BV-03	600	10*3600/490*60	11/0.1	12.5	10	50	19.91	26.51	4410/99.77	4509.77	78.4/99.9	178.3
BV-04	600	10*3600/490*60	9/0.1	12.5	10	50	20.09	26.53	9684/200	9884	172.2/200.2	372.4
BV-05	625	10*3600/490*60	7/0.1	12.5	10	50	18.57	34.1	14210/290.93	14500.92	252.6/291.3	543.9

3.2.2 Results

3.2.2.1 General observations

The creep-fatigue crack propagation model embodied in Eq 2.1 assumes that the crack propagation can be separated into a fatigue and creep part. To compare with experiment we need to determine:

- the crack length versus cycles and time;
- the crack growth per cycle, $\frac{da}{dN}$, and preferably separate the fatigue $\left(\frac{da}{dN}\right)_f$ and creep part $\left(\frac{da}{dN}\right)_c$;
- the load-line displacement, ΔV_{tot} , per cycle and the contribution during the hold time, ΔV_{hold} .

The crack growth per cycle in our tests is typically in the range of $0.1 - 1 \mu\text{m}$, and up to $10 \mu\text{m}$ for the final cycles. The E2760 Standards requires that the PD system should be capable of resolving crack extension of at least $\pm 0.1 \text{ mm}$. The crack length in the experiments is typically 20 mm , so the relative crack growth per cycle is therefore of the order $10^{-4} - 10^{-3}$. **Figure 3.2.4** shows the crack length for the test BV-01 estimated directly from the PD signal and the crack length computed by linearizing the PD signal for each cycle. For the 1 h hold time the crack growth is split into three linear segments of 20 minutes each whereas for the 60 s hold time there is only one linear segment. It follows directly that the noise in crack length measurement from the PD signal is about 0.1 mm in this case. This is much larger than the crack growth per cycle and even by linearization we cannot reliably measure the crack growth per cycle; instead it will be estimated by averaging over a sufficient number of cycles corresponding to a total crack growth larger than the noise. Averaging over a large number of cycles also means that it is not possible distinguish the fatigue and creep contribution from the PD measurements between. The crack length should of course increase monotonically with increasing load cycles. Although the loading is globally tensile, the loading at the crack tip will be both tensile and compressive by shake-down, which means that there may be contact between the crack surfaces in the unloading stage, which could result in reduced PD signal [11] and overall crack behaviour.

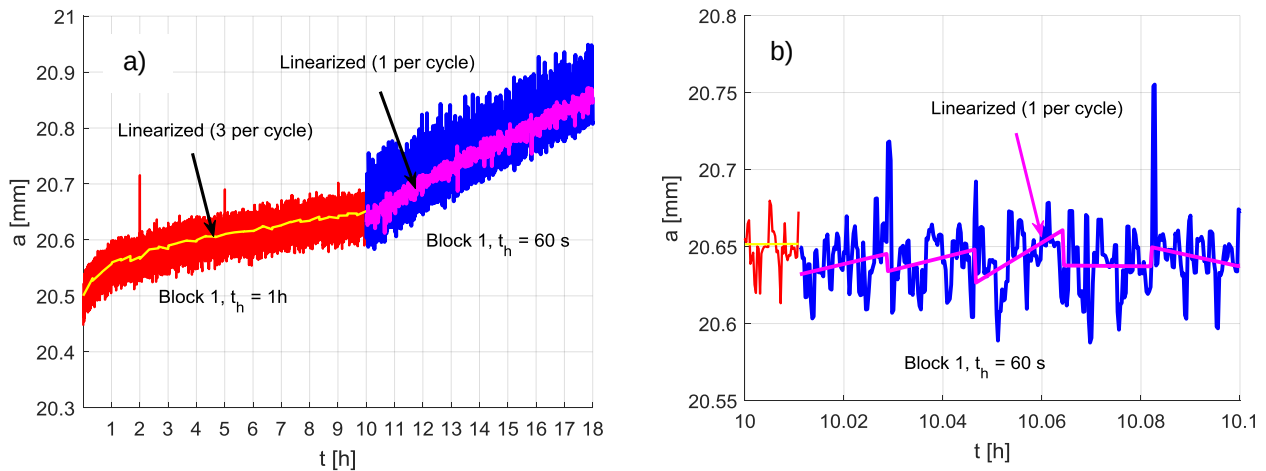


Figure 3.2.4 Crack length versus time for the first two blocks BV-01 (red and blue: directly from PD measurements; yellow and magenta: crack lengths linearized for each cycle) a) $0 - 18 \text{ h}$ b) $10. - 10.1 \text{ h}$.

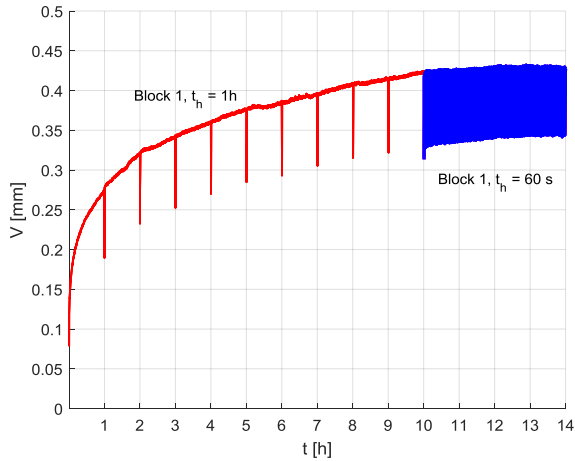


Figure 3.2.5 measured load line displacement versus time for BV-01 for the first 14 hours, short and long hold time

so there is no simple way to extract cycle-by-cycle data for the creep contribution by averaging over several cycles. The accuracy of the load-line displacement differed between tests. **Figure 3.2.7** shows the measured load-line displacement for the test AN-05, and in this case the noise was much smaller and it was possible to determine consistent estimates of the load-line displacement during the hold-time. The load-line displacements used the same procedure and equipment and there is no direct explanation to the differences between the tests.

The load line displacement varies cyclically primarily due to load ramps but also during hold time from accumulated creep damage. In addition to creep the total load line displacement will also increase from accumulated non-reversible plastic deformation (ratcheting).

Figure 3.2.5 depicts the measured load line displacement for BV-01 for the first 14 hours and **Figure 3.2.6** shows the load-line displacement during 30 minutes after 38 hours (the beginning of short hold time cycles for the second block). The load-line displacement from the load range is much larger than load-line displacement during hold time and the noise in the measurement. For the 1 h hold time the load line displacement can be measured with sufficient accuracy whereas for the 60 second hold time there is too much noise for a reliable measurement in this case. The load line displacement varies cyclically with the load, but also depends on the crack length and load history,

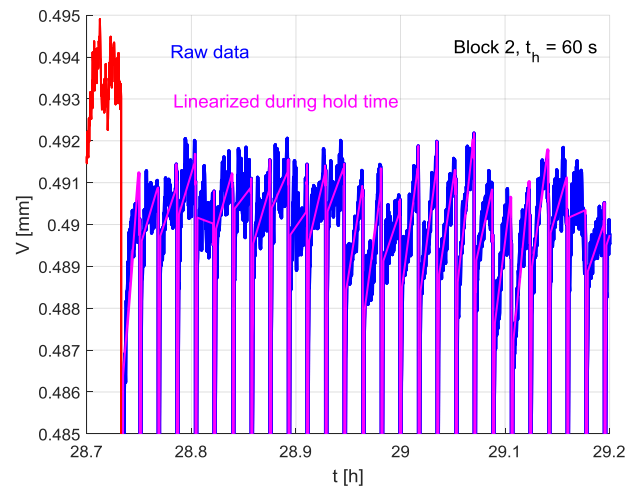
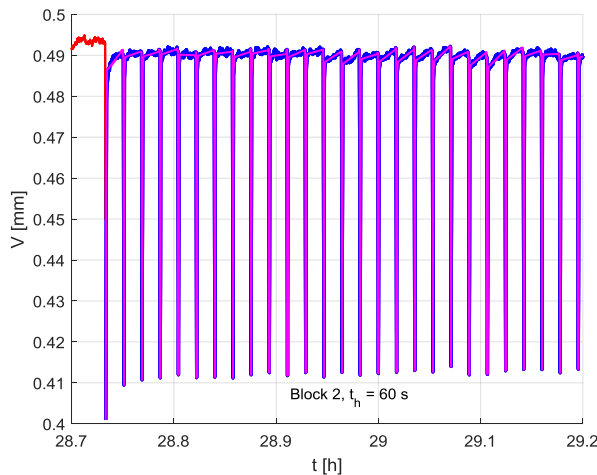


Figure 3.2.6 Measured load line displacement versus time for BV-01 for the second block $t_h = 60$ s

During the 1 h hold time there is sufficient time for a creep zone to extend from the crack tip beyond small scale creep and the creep reversal is minimal. This contrasts with the situation for the 60 second hold time for which creep zone is smaller and there is almost total creep reversal, which is clearly seen for AN-05 but also for BV-01 in spite of the noise. If the load-line displacement during hold time cannot be measured accurately then Eq. 3.4 cannot be used to compute $(C_t)_{avg}$ directly from the measurements. Instead we need to rely on the analytical formula given by Eq. 3.5. The computed value in this case depends strongly on the assumed creep rate. Since the ferritic-martensitic steel does not have a well-defined secondary creep we estimate the creep parameters assuming a constant creep rate from initiation up to rupture from a creep test ($\dot{\epsilon} = A\sigma^n$). This would give a conservative estimate of the creep rate for pure creep. Using typical creep rupture data for P91 by this approach gives $n = 8.11$, $A = 2.28 \cdot 10^{-20}$ for 625°C and $n = 6.91$, $A = 6.57 \cdot 10^{-19}$ for 600°C. **Figure 3.2.8** shows the computed $(C_t)_{avg}$ as function of the crack length using Eq. 3.5. In **Figure 3.2.8a** results are plotted for an applied load P of 9kN for the two temperatures and the three hold times used in the tests: 1h, 600s and

60 seconds. **Figure 3.2.8b** shows the computed values for 625°C and hold time 60 seconds for the three load levels $P = 7, 9$ and 11 kN respectively.

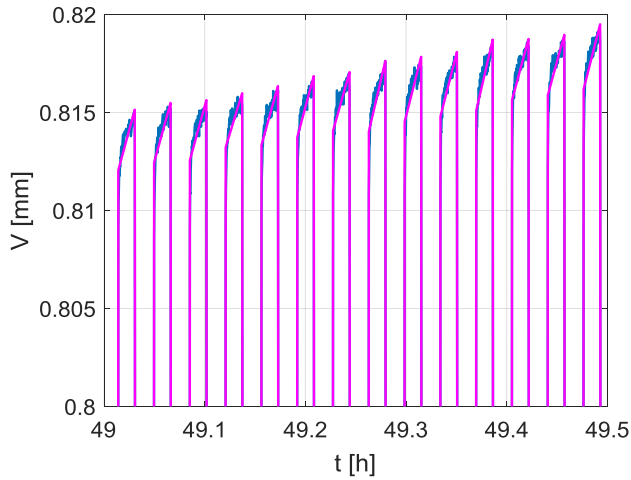


Figure 3.2.7 Measured load line displacement for every second cycle versus time for AN-05 (only every second cycle displayed)

included in the analytical formula. The latter could be quite significant and depends on the number of cycles needed to propagate the crack to a specific crack length.

Although Equation 3.5 is restricted to secondary creep with full creep reversal, it can still give some interesting qualitative results that will be used to interpret some experimental results. Equation 3.5 has two parts: the transient and the steady state (C^*). First of all $(C_t)_{avg}$ increases more than exponentially with increasing crack length. Moreover with increasing crack length the $(C_t)_{avg}$ for different the hold times converge to the same value, which is the steady state value, C^* . The stress intensity factor given by Eq. 3.2, which controls the fatigue part of the crack growth, increases much less as function of the crack length, which means that the creep contribution of the crack growth increases with crack length. **Figure 3.2.8b** indicates that increasing the load from 7 to 11 kN increases the $(C_t)_{avg}$ by a factor 10. The assumption that there is full load reversal leads to an overestimate of the $(C_t)_{avg}$. Primary creep and cyclic softening and associated increased creep rate, which are not included in Eq. 3.5 would increase $(C_t)_{avg}$, but not

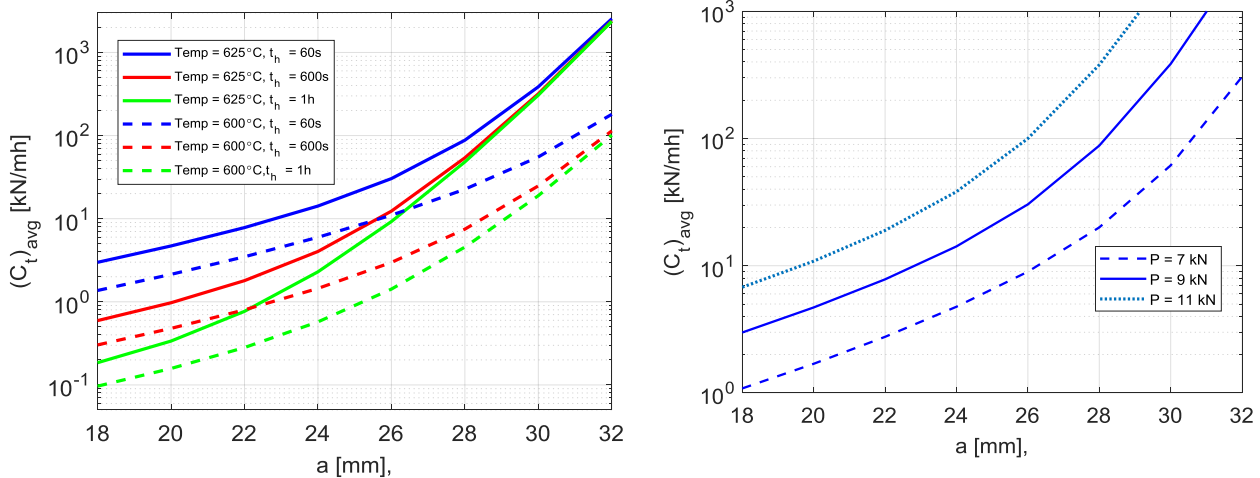


Figure 3.2.8 Computed $(C_t)_{avg}$ versus crack length from Eq. 3.5 a) $P = 9$ kN, $t_h = 60$ s, 600 s and 1 h b) $t_h = 60$ s, $P = 7, 9$ and 11 kN, 625°C: $n = 8.11$, $A = 2.28 \cdot 10^{-20}$ 600°C: $n = 6.91$, $A = 6.57 \cdot 10^{-19}$; creep rate in hours and stress in MPa

3.2.2.2 Constant Hold time: AN03-AN08

Figure 3.2.9a shows the measured crack length versus the number of cycles for the specimen AN03-AN07. The creep effect for a given temperature is seen by the higher crack growth per cycle with increasing hold time and, as expected, the effect is more pronounced at 625°C than at 600°C. Since the initial crack length differs between the tests a direct one-to-one comparison cannot be done to quantify the creep effect. The two tests AN05 and AN07, with nominally the same test conditions, are consistent and the shift between them is due to the difference in the initial crack length. **Figure 3.3.9.b** depicts the measured crack length versus time. The average crack growth rate is clearly reduced by increasing hold time: thus the creep crack growth does not compensate for the reduced number of load cycles.

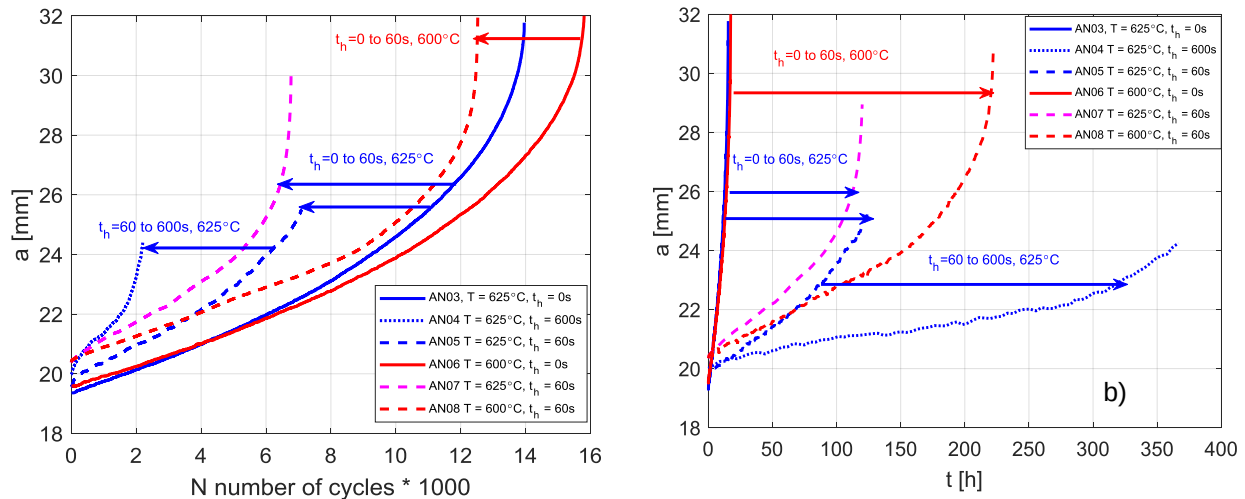


Figure 3.2.9 Measured crack length for the tests AN03-AN08 a) as function of number of cycles b) as function of time.

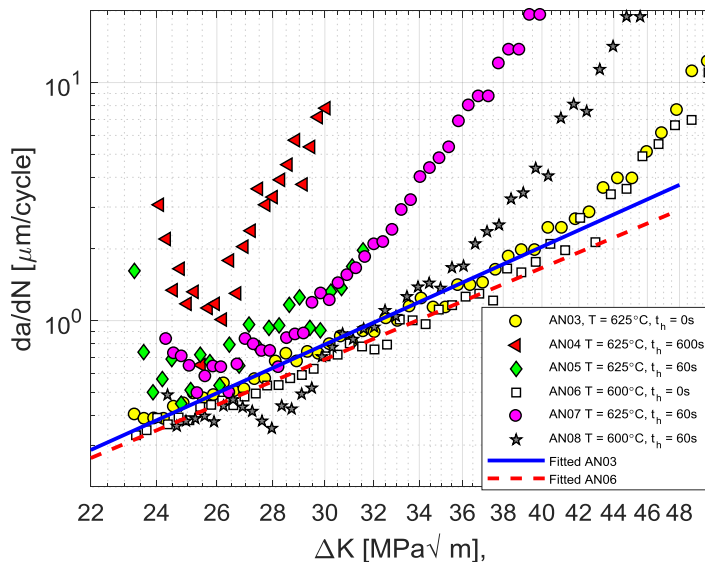


Figure 3.2.10 Computed da/dN versus ΔK from test data and fitted Paris law line for AN03 ($B = 1.11 \cdot 10^{-5}$, $m = 3.287$) and AN06 ($B = 1.96 \cdot 10^{-5}$, $m = 3.076$)

crack growth rate per cycle from the hold time is clearly seen. It can also be noted that for the pure fatigue, the growth rate is only marginally higher for 625°C (AN03) compared to 600°C (AN06). At 600°C , the crack growth rate for 60s hold (AN08) is actually lower than with no hold time (AN06) for ΔK below $30\text{MPa}\sqrt{\text{m}}$, but significantly higher for ΔK above $36\text{MPa}\sqrt{\text{m}}$. There is no reason why hold time should reduce the crack growth per cycle, so this effect is due to scatter in crack length measurements, which is more pronounced for low ΔK values. The increase of the creep effect with increasing ΔK is, however, a direct effect of increased crack length when creep effects become more prominent as discussed above. Note also the characteristic "hook" for the lowest values of ΔK for AN04 with 600s hold time, and to a smaller extent for AN05 and AN07. This is due to the higher creep rate at onset of crack growth when the creep zone at the crack tip is very small and small-scale creep conditions and primary creep prevail. The average creep crack growth rate per cycle during hold time can be estimated from the total crack growth per cycle by subtracting the fatigue part using Paris' law:

A more relevant assessment of the hold-time effect is to compare the crack growth rate per cycle, da/dN , versus the stress intensity factor range, ΔK , where the stress intensity factor range has been computed from Eq. 3.2. **Figure 3.2.10** shows the results for the specimens AN03-AN08. The creep crack growth per cycle is computed by the number of cycles to attain a growth of $50\mu\text{m}$ for AN02, AN05, AN05, AN07 and AN07 and $25\mu\text{m}$ for AN03 and AN05.

This is smaller than the noise in the crack length measurement but used to capture the effect at initial crack growth. The scatter in the measured crack growth rate would be reduced by using larger crack extension. This scatter is larger for the crack growth rates below $1\mu\text{m}$ and differs between the tests; for instance it is larger for AN05 than for AN07. The increase in

$$\frac{da}{dt} = \frac{\frac{da}{dN} B \Delta K^m}{t_h} \quad (3.2.1)$$

Figure 3.2.11a and 3.2.11b show the crack growth rates for the tests AN03 - AN08 versus the number of cycles and crack length respectively. When the crack growth rate is lower with hold than without, as for AN04 for low ΔK , the crack growth rate becomes negative and does not show up in the logarithmic plots. The higher creep rate for the initial cycles is clearly seen and the increase in creep rate for longer crack lengths.

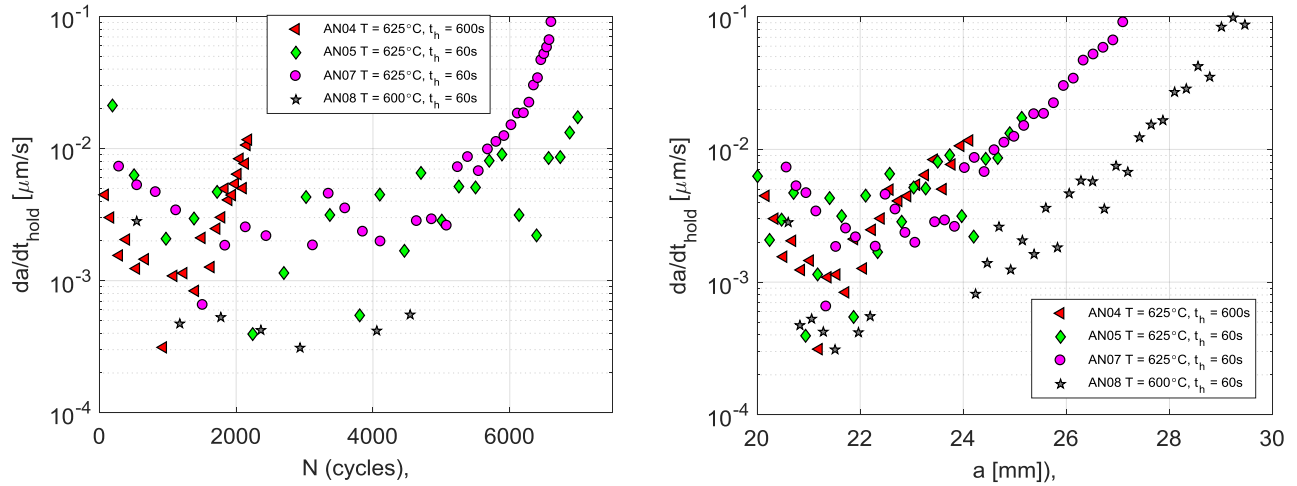


Figure 3.2.11 Calculated average crack growth rate during hold time from experimental data AN4, AN05, AN07, AN08 a) versus number of cycles b) versus crack length.

Rather than using ΔK and Paris law with the parameters B and m being functions of the hold time and temperature, it would be more consistent to associate the creep rate with the $(C_t)_{avg}$. Figure 3.2.12 shows $(C_t)_{avg}$ for the AN04, AN05 and AN07 computed directly from the measurements and Eq. 3.4 and the corresponding values using the analytical expression Eq. 3.5 using the same creep properties as in Figure 3.2.8. For the test AN08 the change in load line displacement during hold time was too small to measure $(C_t)_{avg}$. Both approaches lead to significantly reduced $(C_t)_{avg}$ for 600 seconds hold time compared to 60 seconds hold time. The analytical expression does not account for primary creep and assumes full creep relaxation; the computed $(C_t)_{avg}$ therefore increases monotonically with increasing crack length. It cannot capture the initial drop in $(C_t)_{avg}$, but overestimates $(C_t)_{avg}$ beyond the initial cycles. The key weakness of the analytical expression is that it cannot account for the complex evolution in the cyclic creep properties typical for a ferritic-martensitic steel. As mentioned above the load-line displacement during hold time is a very small fraction of the total load-line displacement in a cycle so there may also be uncertainty for the $(C_t)_{avg}$ derived from the noise in test data.

Figure 3.2.13a depicts the crack growth rate in Figure 3.2.1 versus the $(C_t)_{avg}$ for AN04, AN05, AN07. The solid line is the design data in RCC-MRx at 600°C for "steady state" pure creep ($da/dt = Q \cdot C^q$ with $Q = 0.0061$ and $q = 0.60$ and valid for $0.01 < C^* < 10 \text{ kN/h}$). If the creep crack propagation model is valid then there should be a linear relationship in the loglog-plot. This is reasonably satisfied in this case, but $(C_t)_{avg}$ for the population drop below the linear trend for the lower values. Note that the "hook" is no longer present as the initial crack growth is higher due to the higher $(C_t)_{avg}$. Following the maximum damage approach, Saxena proposed use the maximum of the cyclic and hold time component as a measure of crack growth rate [6,7]:

$$\frac{da}{dt} = \max \left[\left(\frac{da}{dN} \right)_f, \left(\frac{da}{dN} \right)_c \right] / t_{hold} \quad (3.2.2)$$

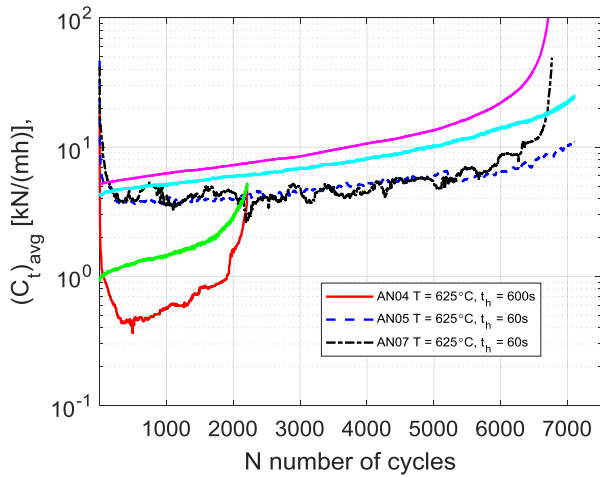


Figure 3.2.12 Computed $(C_t)_{avg}$ versus number of cycles. directly from measurements and Eq. 3.4 and using the analytical formula 3.5

Figure 3.2.13b shows the corresponding crack growth rate versus $(C_t)_{avg}$. For the data points for between 3 and 10 kN/mh the fatigue damage is dominant and there is an overall better linear agreement. The measured crack growth rate is higher RCC-MRx data and this effect increases with increasing $(C_t)_{avg}$. This is not surprising as the high these data points either come from the initial stage where there is scale-scale creep or for the longer crack lengths for which there is more cyclic softening. A similar difference between the dominating damage and hold-time approach was observed in [5,6]

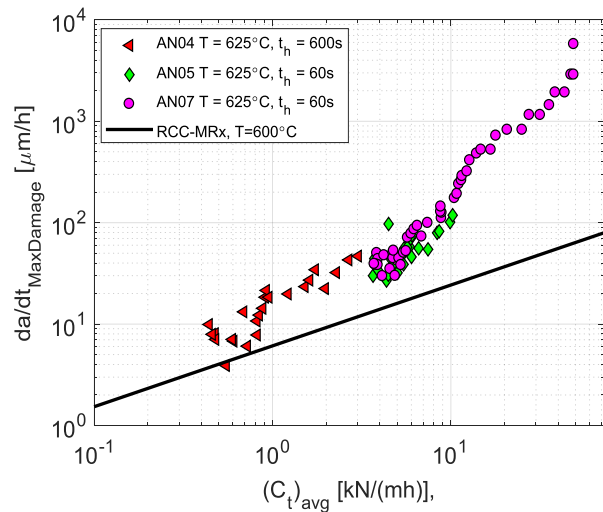
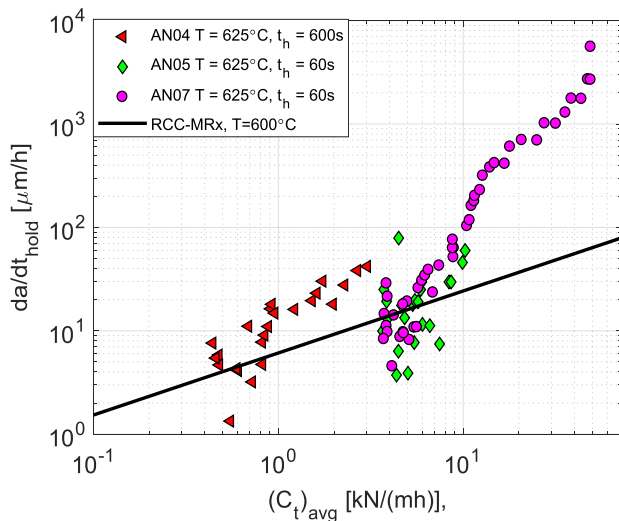


Figure 3.2.13 $(C_t)_{avg}$ versus crack growth rate (taken from Figure 3.11a and 3.12 a) **a)** da/dt_{hold} **b)** $da/dt_{MaxDamage}$

3.2.2.3 Block loading 10·1h+490·60s: BV01-BV05

The measured crack length versus time and number of cycles are shown in **Figure 3.2.14a** and **3.2.14b** respectively. The crack length during the 60 seconds hold times are represented by the blue lines and the 1h hold time by another colour. Up to the final block most of the total crack growth occurs during the 60 seconds hold time blocks for all cases, and this effect is much stronger when the temperature is 600°C, but nevertheless the final cycles were with long hold-time in all cases except for BV04. A plausible explanation is that with increased crack length the creep part increases much more than the fatigue part. The crack growth rate for BV05 is significantly lower due to the lower load and the smaller initial crack length. Up to the final block the crack growth is also much smaller for the 1 hour hold time block than for the 60s hold time block but during the very last long 1 h hold cycle the crack grew unexpectedly from 21 to 34 mm, and a corresponding increase in the load-line displacement was also measured. The test data were checked to verify that the load or temperature had not been accidentally increased. The only physical reason, but still not very plausible, would be that this test had undergone more cycles and therefore would have experienced much more cyclic softening.

Figure 3.2.15 shows the total crack growth per block for the 1h and 60 second hold times respectively. For the 1h hold time the crack growth is initially high and then reaches a minimum but increased significantly for

the final blocks. In **Figure 3.2.4** we also saw that for BV01 crack growth for the first 1 h hold time block decreases with consecutive cycles as the creep zone increases. This pattern is also expected since primary creep effects are initially large and creep crack growth becomes more dominant for longer cracks. This was seen for all tests and this behaviour was also seen for each load block, e.g. the highest crack growth rate was observed for the first cycles in a block.

For the blocks with 60 second the higher crack growth rate for the first block as well as the increase in crack growth for the final blocks were much less pronounced.

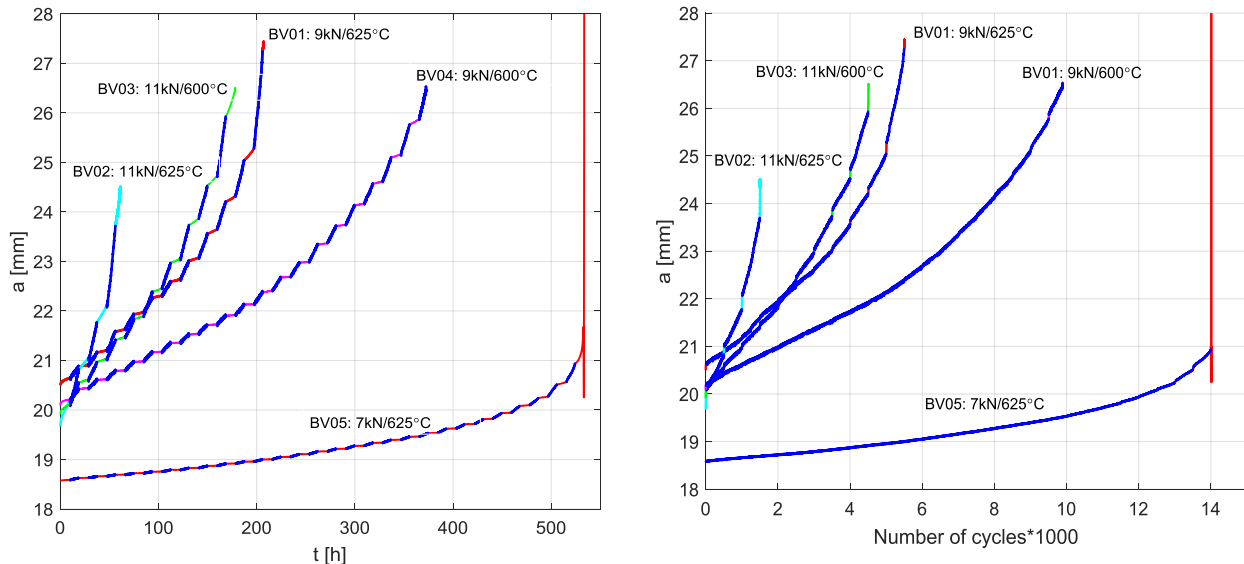


Figure 3.3.14 Measured crack length for the tests BV01-BV05 a) as function of number of cycles b) as function of time.

Figure 3.2.16 shows the computed crack growth per cycle versus the stress intensity factor range together with the fitted curves for pure fatigue. The creep effect is clearly seen for the 1 h hold time (Figure 3.2.16a) and it is as expected higher for 625°C than for 600°C. The crack growth rate is also larger than for AN04 with 600s at comparable ΔK , (e.g. at 30 MPa $\sqrt{m}$ it is ~15mm/cycle for BV01 vs ~8 for AN03), but the difference is smaller than expected. For BV05 with the lowest load, the hold time effect is much larger. In this case to the creep zone at the crack tip is smaller and hence more creep relaxation and primary creep effects are also stronger.

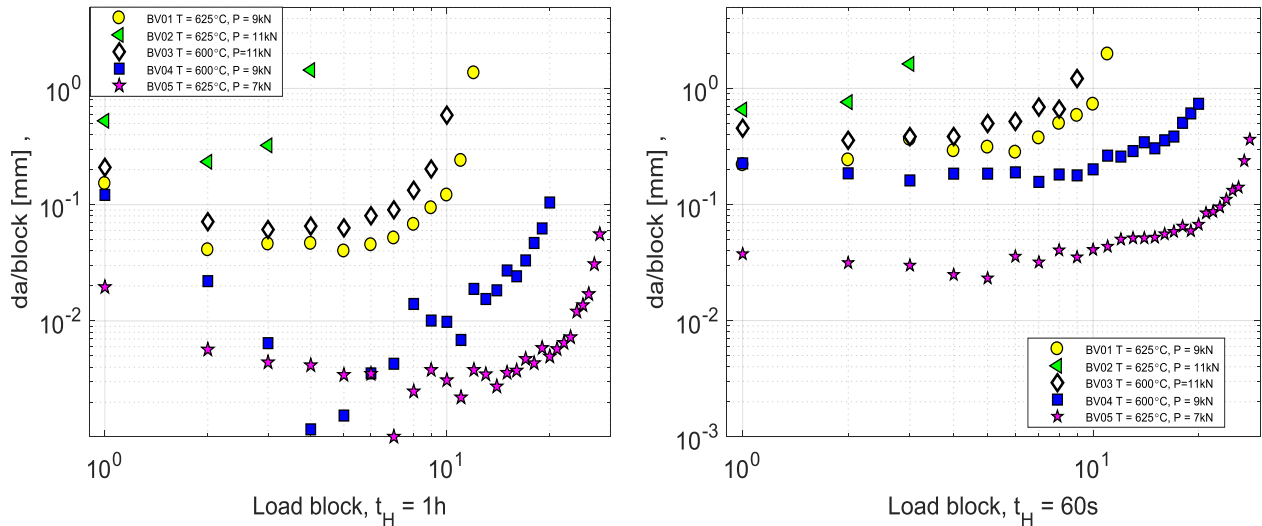


Figure 3.2.15 Measured crack growth per block for the tests BV01-BV05 a) hold time 1h b) hold time 60s

For the 60 seconds hold time (**Figure 3.2.16b**) the creep effect is smaller than what was observed for the constant hold-time tests (**Figure 3.3.10**). This should not be too surprising since the 60s block load is always preceded by the 1h block load and for which extended creep zone has developed. For the short hold time actually the opposite behavior to the 1 h hold-time was observed: the first cycles in the block has slower crack growth as indicated in **Figure 3.2.4**.

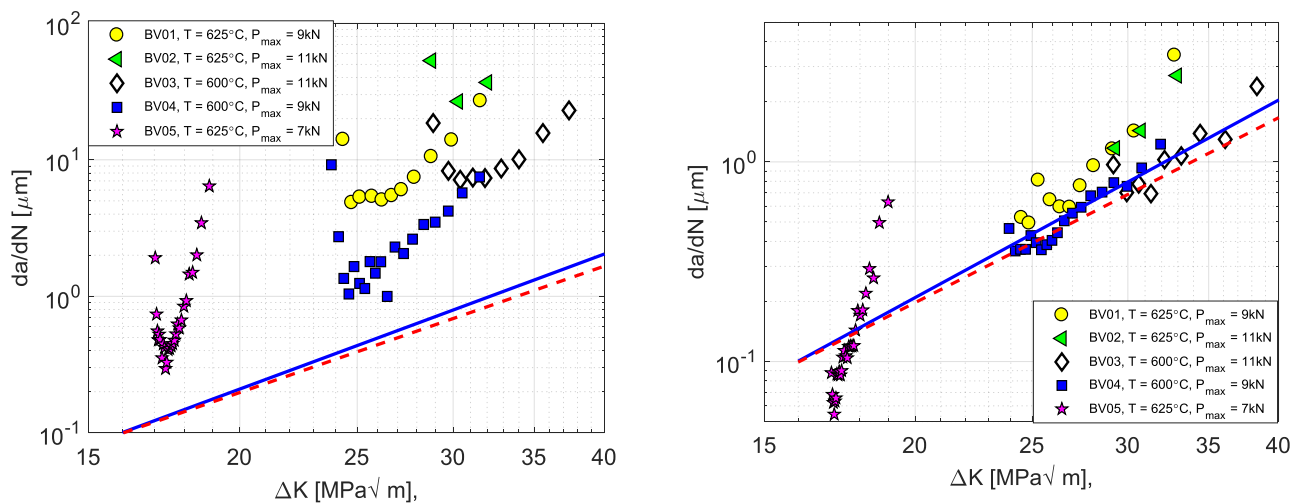


Figure 3.3.16 Computed crack growth rate versus stress intensity factor range for the tests BV01-BV05 a) 1 h hold time b) 60 s hold time

The crack growth rate versus $(C_I)_{avg}$ is depicted in **Figure 3.2.17** for the 1h and 60 second hold-time respectively together with the RCC-MRx design data for steady-state creep [12]. For the 1h hold-time, **Figure 3.2.17a**, $(C_I)_{avg}$ was computed directly from the test data and includes each cycle, whereas for the 60 second hold time the load-displacement curves had too much noise for a reliable estimate and the analytical formula Eq. 3.5 was adopted. For the 1 h hold time the data fall reasonably well on top of each at high crack growth rates but with larger scatter for the lower crack growth rates. The slope is however much steeper than for the RCC-MRx creep data. It can be noted in particular that BV05 with the lower load, now falls in line with the other results. A direct comparison with the corresponding data for the constant hold-time in **Figure 3.2.13** indicate that the crack growth rate is lower for the 1 h hold time than for the tests with constant hold-time.

Figure 3.2.13b and **Figure 3.2.13c** depict results for the 60 s hold time where hold time creep rate in Eq.3.2.1 and maximum damage in Eq. 3.2.2 have employed respectively. For the hold time creep rate the 60 second

hold times have a larger scatter than was observed for the constant hold time and the 1 h hold time with block loading. When the creep rate based on maximum damage is plotted against a much better agreement is achieved (**Figure 3.2.13c**). This is of course expected since both the dominant fatigue crack growth rate, which is computed from Paris law, and the analytical formula for $(C_t)_{avg}$, are both functions of the crack length. There is a very small difference in the Paris law for the two temperatures whereas the computed $(C_t)_{avg}$ is much lower for 600°C than for 625°C due to the difference in the creep parameters A and n . As a consequence the creep rate appears higher for 600°C, which is not logical.

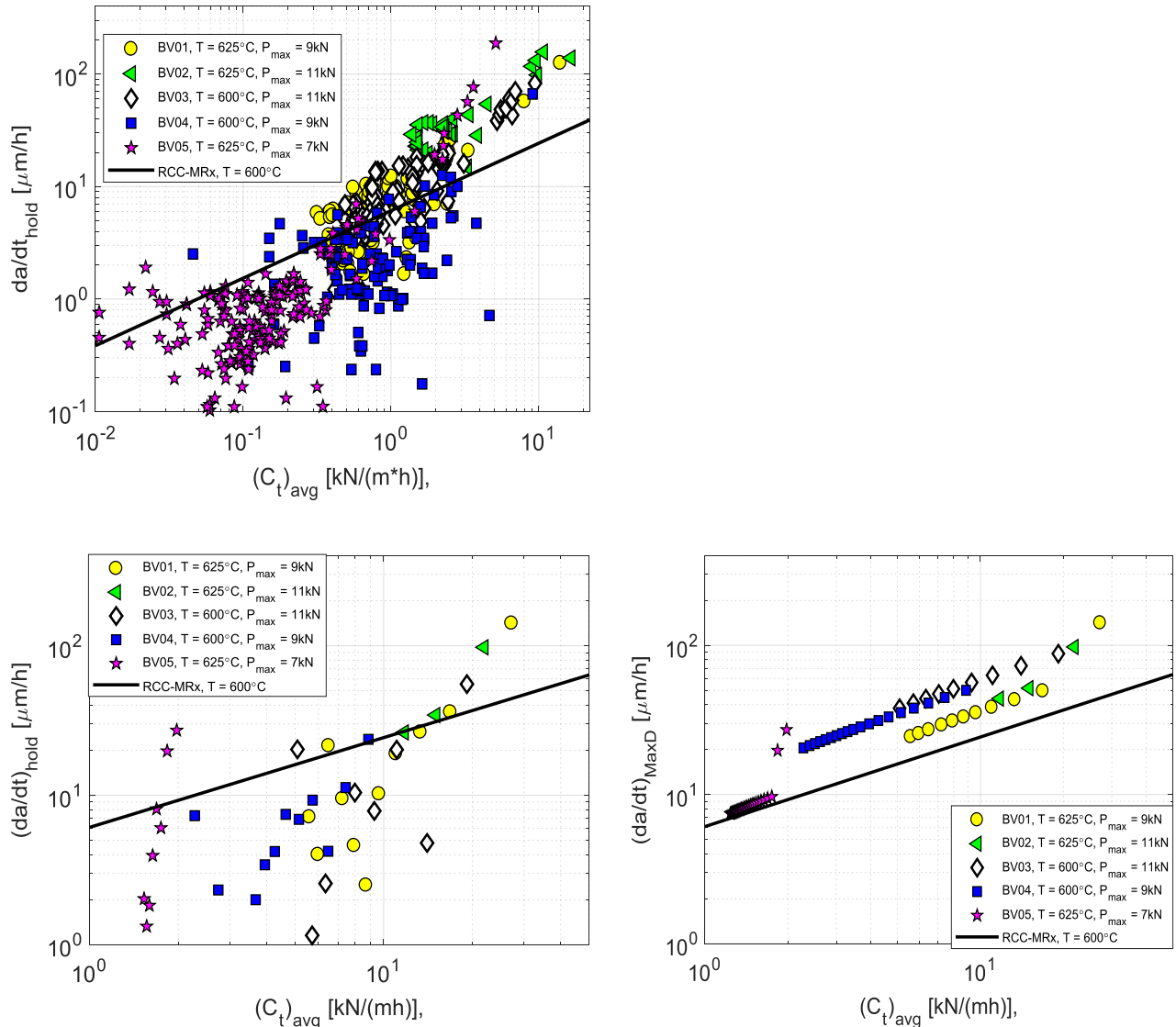


Figure 3.2.17 Crack growth rate versus $(C_t)_{avg}$ a) 1h hold time with $(C_t)_{avg}$ computed directly from test data and Eq. 3.4 b) 60 s hold time with $(C_t)_{avg}$ computed from the analytical formula Eq. 3.5

3.2.3 Concluding remarks JRC tests

The creep-fatigue crack propagation test is a very complex test to perform and requirements need to be followed strictly for valid results. In particular the crack length and the load-line displacement need to be measured continuously during the cycles and with very high accuracy. Difficulties to measure load-line displacement during short hold-times were encountered. In such case the test standard allows the use of closed form solutions that relate the C -integral to load and geometry are based on secondary creep with no cyclic hardening or softening. Such models may give good accuracy for austenitic steels but for ferritic-martensitic steels, which has no secondary creep and strong cyclic softening with associated increase in creep rates, the applicability becomes limited.

The present set of tests clearly demonstrated the hold time effect for creep-fatigue crack propagation and allowed reasonable quantification. Additional tests and further improved procedures are needed to better quantify creep-fatigue crack propagation. The dominant damage approach to determine the creep rate gives a better linear relationship than the more natural approach where the creep rate is computed from the hold time. As regards prediction of creep-fatigue crack propagation by models it is necessary to describe the cyclic visco-plastic properties. For ferritic-martensitic steels this is a particularly difficult.

3.3 CIEMAT Creep-fatigue tests

3.3.1 Introduction

Creep-fatigue crack growth (CFCG) rate of candidate materials for Gen IV reactors are needed due to the observed decrease of fatigue strength after hold time periods. Due to the importance of generating reliable data, ASTM developed a standard entitled, "E-2760-10: Standard Test Method for Creep-fatigue Crack Growth Testing" published in 2010. A round robin test was launched by ASTM involving participants from North America, Europe and Asia, is to commence soon. The round-robin was conducted using standard C(T) specimens using test material P91 steel. The primary objective of the round-robin was to conduct testing to further validate the standard and to assess precision and bias in the data generated while using the procedures specified in this ASTM Standard. CIEMAT participated in this round robin tests by testing two specimens at 625°C, hold time of 600 s and two different loads range: 9-0.9 kN and 7.5-0.57 kN.

In this report, these RR tests are presented as well as the dedicated CFCG performed within MATISSE that are summarized.

3.3.2 Materials

The materials tested were three different heats of T91 F/M steels: P91-RR, that is the materials used for the ASTM RR; T91-D that is a T91 heat from DEMETRA project; and T91-JRC supplied by JRC for the MATISSE project. A heat of 316L from DEMETRA project is also included.

The P91-RR was supplied by EPRI and was obtained from an ex-service pipe section with an outer diameter of 480 mm and a wall thickness of 45 mm from which specimens blanks were machined. The material was re-heat-treated to ensure consistency with original microstructure.

The T91-D and 316L-D were supplied by Industel within DEMETRA FP7 project. The first is the ferritic-martensitic steel T91 (Grade 91 Class 2 / S50460), according to ASTM standard A387-Ed99. The second material is the austenitic stainless steel 316L (S31603), according to ASTM A240-Ed02. Both materials were delivered as hot rolled and heat treated plates with a thickness of 15 mm by Industeel. For T91, the normalizing treatment consisted of heating the alloy to 1050°C; holding for 1 minute per millimeter, thus 15 minutes and then cooling with water to room temperature. This treatment produced a fully martensitic structure. The tempering treatment consisted of heating the normalized steel to 770°C, holding for three minutes per millimeter of thickness, thus 45 minutes and then air cooling in still air to room temperature. The 316L was solution annealed at 1050°C - 1100°C¹.

3.3.3 Test specimen

Compact tension specimen, CT, with a width, $W = 50$ mm, thickness, $B = 12.5$ mm and initial notch, $a_0 = 13.5$ mm where used. Reduced thickness was 12.5 mm, see **Figure 3.3.1**. All the specimens were mechanised in the TL orientation. The specimens were pre-cracked to an initial crack length to width ratio, a/W , of about 0.4, under cyclic loading at room temperature. The specimens were then side-grooved (10% of thickness on each side with 60 degree V-notch) to prevent tunneling at elevated temperature.

¹ EUROTRANS FI6W-CT-2004-516520 Deliverables: D 4.5a & D 4.6 July, 2005

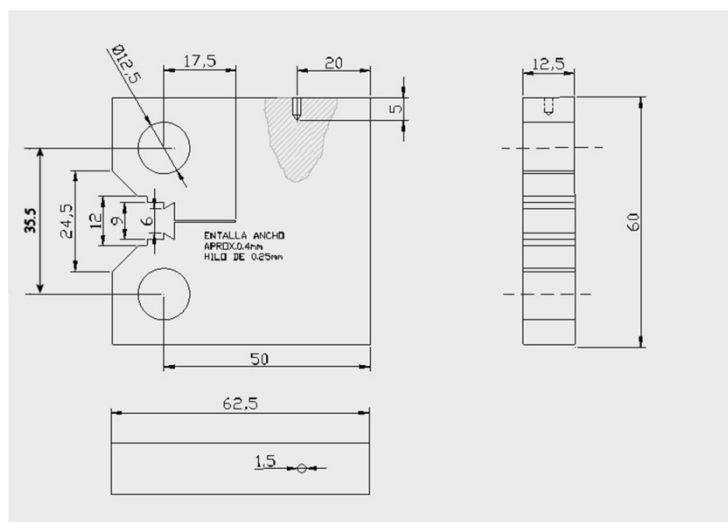


Figure 3.3.1 Compact tension specimens used

3.3.4 Experiments and test procedure

All the tests were conducted in a servo-hydraulic test machine. Potential drop technique was used to monitor crack length during the tests. Load-line displacement were measured with a high temperature extensometer located at the load line, see **Figure 3.3.2**. Temperature were measured at the specimen surface.

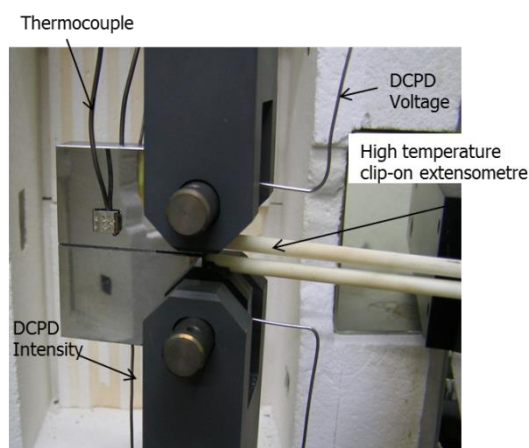


Figure 3.3.2. Experimental set-up of CIEMAT tests

Table 3.3.1. CIEMAT CFCG tests matrix

Material	Specimen ID	Temperature (°C)	Maximum Load (kN)	Hold time (s)
T91-RR	3.1.27	625	9	600
	3.1.28	625	7.5	600
T91-D	D2	625	9	60
	D3	625	9	600
	D4	625	7.5	60
T91-JRC	J1	600	9	600
	J2	625	9	60
	J3	600	7.5	60
316L-D	3D1	625	9	60

The waveforms for loading and unloading portions were triangular and the loading/unloading times were held constant (2 s rise and decay times). Hold times of predetermined duration (60 and 600 s) were superimposed on the triangular waveforms at maximum load. Tests were carried out at 625°C and 600°C at a load ratio of $R=0.1$. Test matrix can be seen in **Table 3.3.1**

3.3.5 Analysis Procedure

The following signals are recorded per cycle:

- Load (kN) = Load;
- DCPD (V) = Potential drop signal;
- VLL (mm) = Load line displacement measured with the extensometer at the load line;
- Time (sec) = Time.

The crack length was monitored by the potential drop technique. After the tests, the final crack length a_f was measured by averaging 5 equi-distance measurements and the measured crack length is adjusted by:

$$a_{adjusted}(i) = a_0 + (a_f - a_0) \frac{(DCPD_{avg}(i) - DCPD_{final})}{(a_f - a_0)} \text{ where}$$

$$DCPD_{avg}(i) = \frac{DCPD_{max}(i) - DCPD_{min}(i)}{2}$$

The crack-tip parameter $(C_t)_{avg}$ is calculated as follows:

$$(C_t)_{avg} = \frac{\Delta P \Delta V_c}{\sqrt{B_N B W t_h}} \frac{F'}{F} \quad [MJ/m^2-h]$$

Where ΔV_c is the difference in force-line displacement between the end and start of the hold time, during a cycle and F'/F is given by:

$$\frac{F'}{F} = \left[\left(\frac{1}{2+a/W} \right) + \left(\frac{3}{2(1-a/W)} \right) \right] + \left[\frac{(4.64 - 26.64(a/W) + 44.16(a/W)^2) - 22.4(a/W)^3}{0.886 + 4.64(a/W) - 13.32(a/W)^2 + 14.72(a/W)^3 - 5.6(a/W)^4} \right]$$

For each cycle (i) the difference in force line displacement between the end and start of the hold time ΔV_c is calculated and later on smoothed to facilitate further calculations.

As it is recommended by Saxena et al², in order to determine if the creep-fatigue crack growth rates should be expressed as da/dN versus ΔK for a fixed hold time in the case of creep-brittle materials or for ductile creep-fatigue they should be expressed as $(da/dt)_{avg}$ versus $(C_t)_{avg}$. The total measured change in the force-line displacement rate, ΔV , can be separated into an instantaneous elastic part, ΔV_e , and a time-dependent part that is associated with the accumulation of creep strains, ΔV_c . ΔV_e is estimated using the following equation.

$$\Delta V_e = \frac{t_h \left(\frac{da}{dt} \right)_{avg}}{P} B \left[\frac{2\Delta K^2}{E'} \right]$$

² S.B. Narasimhachary, A. Saxena / International Journal of Fatigue 56 (2013) 106–113

Following Saxena's recommendations, the material can be classified as creep-brittle if $\Delta V_e \geq 0.5 \Delta V$ and the crack growth behavior may be characterized by the stress-intensity factor. On the other hand, if $\Delta V_e < 0.5 \Delta V$, creep-ductile conditions are said to prevail and the crack growth behavior should be characterized by $(C_t)_{avg}$. Crack growth rate has been calculated for crack increments of 0.25 mm.

3.3.6 Results

Material T91-RR

An example of the fitting procedure of the load-line displacement difference at each hold time cycle is seen at Figure 3.3.3, where the original signal of the first 5 cycles of the load-line displacement is seen.

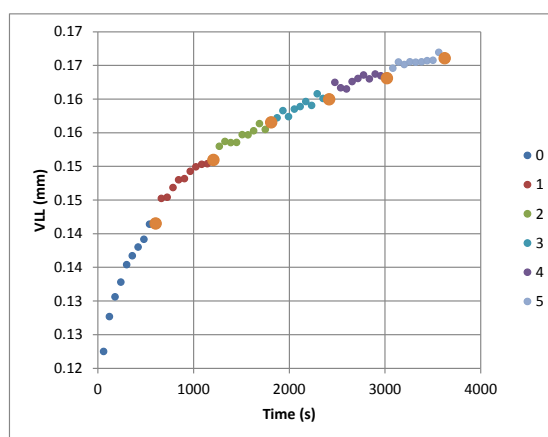


Figure 3.3.3. Load line displacement for the first 5 cycles. Specimen 3.1.28

Figure 3.3.4(a) shows the calculated load-line difference and the smoothed signal and Figure 3.3.4(b) shows the smoothed signal with the reduced data used for the calculation of $(C_t)_{avg}$.

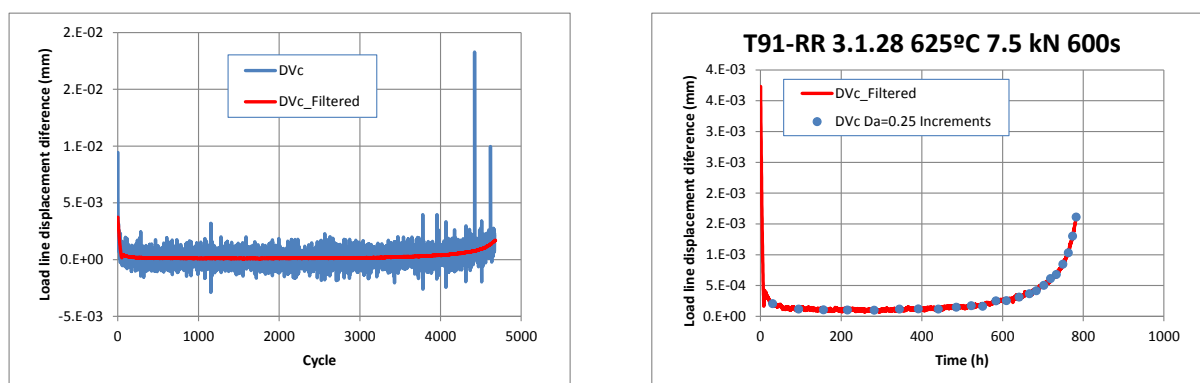


Figure 3.3.4. Load line displacement difference. Specimen 3.1.28

The results of the crack growth da/dN and da/dt with the creep-fatigue parameter $(C_t)_{avg}$ and the DK value can be seen in Table 3.3.2 and Table 3.3.3 for specimens 3.1.28 and 3.1.27 respectively.

Table 3.3.2: Creep-crack growth results. Material T91-RR, specimen 3.1.28. 625°C, 7.5 kN, hold 600 s

N	a	t	ΔV_c	ΔV_e	ΔK	da/dN	Ct	da/dt
Cycle	mm	h	mm	mm	MPa $\sqrt{m}$	mm/cycle	MJ/(m ² h)	mm/h
	20.65							
180	20.19	30.12	2.04E-04		19.64			
564	20.46	94.55	1.16E-04		19.93			
932	20.66	156.30	1.04E-04	2.83E-06	20.15	4.76E-04	2.22E-05	2.85E-03
1287	20.96	215.86	1.01E-04	4.22E-06	20.47	6.87E-04	2.16E-05	4.12E-03
1680	21.21	281.80	9.97E-05	4.56E-06	20.75	7.23E-04	2.15E-05	4.34E-03
2053	21.42	344.39	1.16E-04	5.12E-06	21.00	7.92E-04	2.51E-05	4.75E-03
2334	21.70	391.54	1.20E-04	5.69E-06	21.32	8.55E-04	2.64E-05	5.13E-03
2630	21.95	441.20	1.16E-04	6.70E-06	21.62	9.79E-04	2.55E-05	5.87E-03
2893	22.21	485.33	1.46E-04	7.80E-06	21.94	1.11E-03	3.25E-05	6.64E-03
3115	22.44	522.58	1.70E-04	8.88E-06	22.22	1.23E-03	3.81E-05	7.36E-03
3284	22.69	550.94	1.63E-04	9.81E-06	22.54	1.32E-03	3.68E-05	7.91E-03
3479	22.95	583.66	2.51E-04	1.09E-05	22.88	1.42E-03	5.71E-05	8.54E-03
3634	23.19	609.66	2.55E-04	1.20E-05	23.19	1.52E-03	5.84E-05	9.13E-03
3821	23.46	641.04	3.12E-04	1.39E-05	23.57	1.71E-03	7.23E-05	1.03E-02
3973	23.71	666.54	3.64E-04	1.66E-05	23.91	1.98E-03	8.51E-05	1.19E-02
4076	23.91	683.83	4.14E-04	1.89E-05	24.20	2.20E-03	9.74E-05	1.32E-02
4187	24.19	702.45	5.03E-04	2.17E-05	24.60	2.45E-03	1.20E-04	1.47E-02
4286	24.45	719.06	6.11E-04	2.44E-05	24.99	2.67E-03	1.47E-04	1.60E-02
4372	24.69	733.49	6.80E-04	2.67E-05	25.35	2.83E-03	1.65E-04	1.70E-02
4469	24.96	749.77	8.45E-04	3.12E-05	25.79	3.21E-03	2.07E-04	1.92E-02
4549	25.21	763.19	1.03E-03		26.19			
4612	25.41	773.76	1.30E-03		26.52			
4666	25.71	782.82	1.61E-03		27.02			

Table 3.3.3: Creep-crack growth results. Material T91-RR, specimen 3.1.27. 625°C, 9 kN, hold 600 s

N	a	t	ΔV_c	ΔV_e	ΔK	da/dN	(Ct)avg	da/dt
Cycle	mm	h	mm	mm	MPa $\sqrt{m}$	mm/cycle	MJ/(m ² h)	mm/h
	20.65							
20	20.47	3.27	2.25E-03		24.24			
149	20.76	24.92	4.48E-04		24.61			
238	21.06	39.85	4.01E-04	1.53E-05	25.02	2.02E-03	1.05E-04	1.21E-02
365	21.28	61.16	3.72E-04	1.70E-05	25.33	2.19E-03	9.79E-05	1.32E-02
505	21.55	84.65	3.03E-04	1.74E-05	25.70	2.17E-03	8.03E-05	1.30E-02
629	21.81	105.46	3.23E-04	1.93E-05	26.07	2.35E-03	8.62E-05	1.41E-02
733	22.06	122.91	3.82E-04	2.24E-05	26.44	2.65E-03	1.03E-04	1.59E-02
819	22.28	137.34	4.50E-04	2.59E-05	26.77	2.99E-03	1.22E-04	1.79E-02
902	22.55	151.26	5.09E-04	2.88E-05	27.18	3.22E-03	1.39E-04	1.93E-02
972	22.79	163.01	4.59E-04	3.14E-05	27.55	3.42E-03	1.26E-04	2.05E-02
1028	23.01	172.40	6.25E-04	3.43E-05	27.91	3.64E-03	1.73E-04	2.18E-02
1115	23.30	187.00	6.62E-04	3.84E-05	28.37	3.94E-03	1.85E-04	2.36E-02
1179	23.55	197.74	7.61E-04	4.40E-05	28.80	4.38E-03	2.15E-04	2.63E-02
1234	23.81	206.97	7.37E-04	5.04E-05	29.23	4.87E-03	2.10E-04	2.92E-02
1281	24.03	214.85	8.42E-04	5.83E-05	29.62	5.50E-03	2.42E-04	3.30E-02
1326	24.30	222.41	1.36E-03	6.94E-05	30.10	6.33E-03	3.95E-04	3.80E-02
1369	24.56	229.62	1.70E-03	8.57E-05	30.57	7.58E-03	4.98E-04	4.55E-02
1399	24.80	234.65	1.99E-03	1.01E-04	31.02	8.67E-03	5.89E-04	5.20E-02
1423	25.02	238.68	2.31E-03	1.13E-04	31.45	9.47E-03	6.90E-04	5.68E-02
1450	25.30	243.21	2.78E-03	1.28E-04	32.01	1.03E-02	8.40E-04	6.19E-02
1472	25.54	246.90	3.29E-03	1.40E-04	32.50	1.10E-02	1.00E-03	6.59E-02
1494	25.78	250.59	3.91E-03	1.62E-04	33.00	1.23E-02	1.21E-03	7.39E-02
1518	26.05	254.62	4.89E-03	2.04E-04	33.58	1.50E-02	1.53E-03	8.98E-02
1537	26.31	257.81	6.65E-03	2.73E-04	34.14	1.94E-02	2.10E-03	1.16E-01
1547	26.56	259.49	8.02E-03		34.71			
1554	26.74	260.66	9.13E-03		35.12			
1566	27.04	262.67	1.13E-02		35.84			

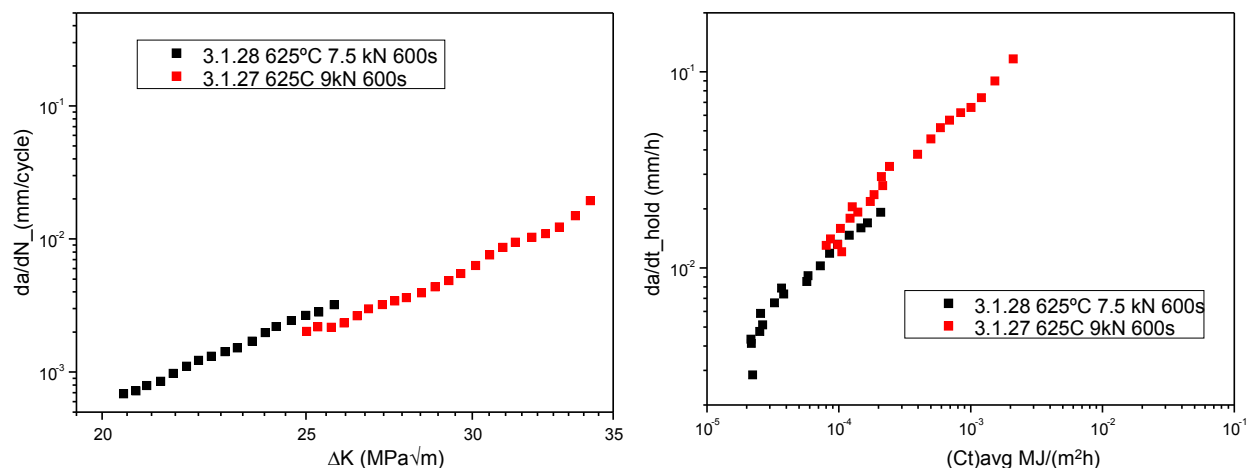


Figure 3.3.5. Creep-fatigue crack growth data for T91-RR material

As can be seen in **Figure 3.3.5**, the crack growth data are not dependent on the load used for the tests. In this figure the typical hook-type at the beginning of the test, when crack growth is represented vs ΔK , is not seen mainly due to the reducing technique. If lower crack growth increments are used, e.g. 0.15 mm increment, the hook-type is clearly seen.

Material T91-D

In this case, the tests have been performed varying the load and the hold time. The results of the three specimens can be seen in the following tables:

Table 3.3.4: Creep-crack growth results. Material T91-D, specimen D2. 625°C, 9 kN, hold 60 s

N	a	t	ΔVc	ΔVe	ΔK	da/dN	Ct	da/dt
Cycle	mm	h	mm	mm	MPa $\sqrt{m}$	mm/cycle	MJ/(m ² h)	mm/h
	20.65							
339	20.56	6.02	1.46E-04		24.55			
849	20.81	15.09	7.26E-05		24.88			
1456	21.00	25.89	6.91E-05	3.00E-06	25.13	3.99E-04	1.82E-04	2.39E-02
1982	21.30	35.25	5.85E-05	3.58E-06	25.55	4.60E-04	1.55E-04	2.76E-02
2528	21.55	44.96	7.17E-05	3.83E-06	25.90	4.78E-04	1.92E-04	2.87E-02
3044	21.78	54.14	6.96E-05	4.15E-06	26.22	5.06E-04	1.87E-04	3.04E-02
3584	22.06	63.74	6.73E-05	4.39E-06	26.63	5.19E-04	1.82E-04	3.12E-02
4016	22.29	71.43	9.44E-05	4.74E-06	26.99	5.46E-04	2.58E-04	3.28E-02
4489	22.56	79.84	1.09E-04	5.26E-06	27.39	5.88E-04	2.99E-04	3.53E-02
4917	22.80	87.45	1.17E-04	5.73E-06	27.76	6.23E-04	3.24E-04	3.74E-02
5310	23.05	94.44	1.28E-04	6.25E-06	28.16	6.61E-04	3.59E-04	3.97E-02
5673	23.32	100.90	1.49E-04	6.81E-06	28.59	6.99E-04	4.21E-04	4.19E-02
6037	23.57	107.37	1.58E-04	7.42E-06	29.01	7.39E-04	4.49E-04	4.43E-02
6370	23.82	113.30	2.10E-04	8.09E-06	29.44	7.83E-04	6.03E-04	4.70E-02
6644	24.03	118.17	2.40E-04	8.87E-06	29.81	8.37E-04	6.95E-04	5.02E-02
6965	24.30	123.88	3.04E-04	9.87E-06	30.29	9.03E-04	8.88E-04	5.42E-02
7200	24.52	128.06	3.12E-04	1.09E-05	30.70	9.74E-04	9.20E-04	5.84E-02
7476	24.81	132.97	4.02E-04	1.24E-05	31.24	1.07E-03	1.20E-03	6.39E-02
7734	25.06	137.56	5.21E-04	1.42E-05	31.73	1.18E-03	1.57E-03	7.10E-02
7923	25.32	140.92	6.96E-04	1.62E-05	32.23	1.31E-03	2.12E-03	7.85E-02
8117	25.56	144.37	7.93E-04	1.86E-05	32.71	1.46E-03	2.44E-03	8.75E-02
8279	25.81	147.25	1.01E-03	2.09E-05	33.24	1.59E-03	3.14E-03	9.52E-02
8423	26.06	149.82	1.17E-03	2.38E-05	33.78	1.75E-03	3.67E-03	1.05E-01
8567	26.31	152.38	1.54E-03	2.77E-05	34.33	1.97E-03	4.89E-03	1.18E-01
8693	26.54	154.62	2.00E-03	3.21E-05	34.84	2.22E-03	6.42E-03	1.33E-01
8808	26.81	156.66	2.53E-03	3.95E-05	35.48	2.63E-03	8.22E-03	1.58E-01
8895	27.05	158.21	3.17E-03		36.05			
8980	27.29	159.72	4.12E-03		36.64			
9041	27.56	160.81	5.30E-03		37.31			

Table 3.3.5: Creep-crack growth results. Material T91-D, specimen D3. 625°C, 9 kN, hold 600 s

N	a	t	ΔV_c	ΔV_e	ΔK	da/dN_poly	Ct	da/dt_hold
Cycle	mm	h	mm	mm	MPa√m	mm/cycle	MJ/(m ² h)	mm/h
	20.65							
165	20.88	27.60	5.13E-04		24.83			
406	21.14	68.04	3.58E-04		25.18			
781	21.40	130.96	2.37E-04	6.85E-06	25.53	8.68E-04	6.27E-05	5.21E-03
1056	21.64	177.10	2.61E-04	6.82E-06	25.88	8.42E-04	6.95E-05	5.05E-03
1349	21.89	226.26	2.98E-04	7.48E-06	26.23	8.98E-04	7.98E-05	5.39E-03
1636	22.11	274.42	3.39E-04	8.94E-06	26.55	1.05E-03	9.14E-05	6.29E-03
1909	22.39	320.23	4.21E-04	1.12E-05	26.97	1.27E-03	1.14E-04	7.61E-03
2074	22.62	347.91	4.65E-04	1.22E-05	27.32	1.35E-03	1.28E-04	8.08E-03
2237	22.89	375.26	6.82E-04	1.40E-05	27.74	1.51E-03	1.89E-04	9.04E-03
2390	23.14	400.93	7.49E-04	1.62E-05	28.13	1.69E-03	2.09E-04	1.01E-02
2567	23.36	430.63	1.04E-03	1.95E-05	28.49	1.98E-03	2.92E-04	1.19E-02
2688	23.64	450.94	1.27E-03	2.31E-05	28.96	2.27E-03	3.60E-04	1.36E-02
2785	23.89	467.21	1.60E-03	2.74E-05	29.38	2.62E-03	4.57E-04	1.57E-02
2879	24.15	482.98	1.76E-03	3.38E-05	29.83	3.14E-03	5.08E-04	1.89E-02
2956	24.35	495.90	1.92E-03	4.18E-05	30.20	3.78E-03	5.59E-04	2.27E-02
3024	24.61	507.31	3.68E-03	5.42E-05	30.67	4.76E-03	1.08E-03	2.86E-02
3077	24.89	516.21	4.58E-03		31.20			
3119	25.14	523.25	5.90E-03		31.67			
3150	25.38	528.45	7.77E-03		32.16			

Table 3.3.6: Creep-crack growth results. Material T91-D, specimen D4. 625°C, 7.5 kN, hold 600 s

N	a	t	ΔV_c	ΔV_e	ΔK	da/dN_poly	Ct	da/dt_hold
Cycle	mm	h	mm	mm	MPa $\sqrt{m}$	mm/cycle	MJ/(m ² h)	mm/h
	20.65							
427	20.50	71.56	1.20E-04		20.38			
1316	20.76	220.73	3.27E-05		20.66			
2261	20.98	379.29	6.45E-05	1.84E-06	20.91	2.93E-04	1.42E-05	1.76E-03
2919	21.24	489.69	8.60E-05	2.28E-06	21.20	3.54E-04	1.90E-05	2.12E-03
3633	21.50	609.49	6.69E-05	2.66E-06	21.51	4.01E-04	1.49E-05	2.41E-03
4305	21.75	722.25	8.06E-05	3.13E-06	21.80	4.59E-04	1.80E-05	2.76E-03
4811	22.00	807.15	1.15E-05	3.63E-06	22.11	5.19E-04	2.60E-06	3.11E-03
5228	22.23	877.12	1.25E-04	4.08E-06	22.39	5.68E-04	2.84E-05	3.41E-03
5673	22.49	951.79	1.14E-04	4.60E-06	22.72	6.22E-04	2.61E-05	3.73E-03
6050	22.75	1015.04	1.52E-04	5.23E-06	23.06	6.86E-04	3.51E-05	4.12E-03
6433	23.00	1079.31	1.43E-04	5.96E-06	23.38	7.60E-04	3.32E-05	4.56E-03
6735	23.22	1129.98	1.73E-04	6.64E-06	23.67	8.27E-04	4.06E-05	4.96E-03
7034	23.51	1180.15	1.78E-04	7.63E-06	24.07	9.19E-04	4.21E-05	5.52E-03
7301	23.75	1224.95	2.36E-04	8.93E-06	24.42	1.05E-03	5.61E-05	6.27E-03
7541	23.99	1265.22	3.57E-04	1.04E-05	24.76	1.19E-03	8.59E-05	7.13E-03
7740	24.23	1298.61	3.29E-04	1.17E-05	25.12	1.30E-03	7.97E-05	7.79E-03
7918	24.49	1328.47	5.12E-04	1.35E-05	25.52	1.44E-03	1.25E-04	8.66E-03
8077	24.74	1355.15	5.67E-04	1.52E-05	25.90	1.58E-03	1.40E-04	9.47E-03
8254	24.99	1384.85	7.73E-04	1.76E-05	26.30	1.77E-03	1.93E-04	1.06E-02
8387	25.26	1407.17	7.37E-04	2.02E-05	26.73	1.97E-03	1.86E-04	1.18E-02
8511	25.49	1427.97	1.19E-03	2.35E-05	27.12	2.23E-03	3.04E-04	1.34E-02
8618	25.75	1445.93	1.48E-03	2.77E-05	27.57	2.54E-03	3.82E-04	1.52E-02
8716	25.99	1462.37	1.97E-03	3.29E-05	28.00	2.93E-03	5.14E-04	1.76E-02
8803	26.25	1476.97	2.62E-03	4.04E-05	28.47	3.48E-03	6.91E-04	2.09E-02
8874	26.50	1488.88	3.08E-03		28.94			
8935	26.76	1499.11	4.27E-03		29.43			
8981	27.01	1506.83	5.31E-03		29.92			

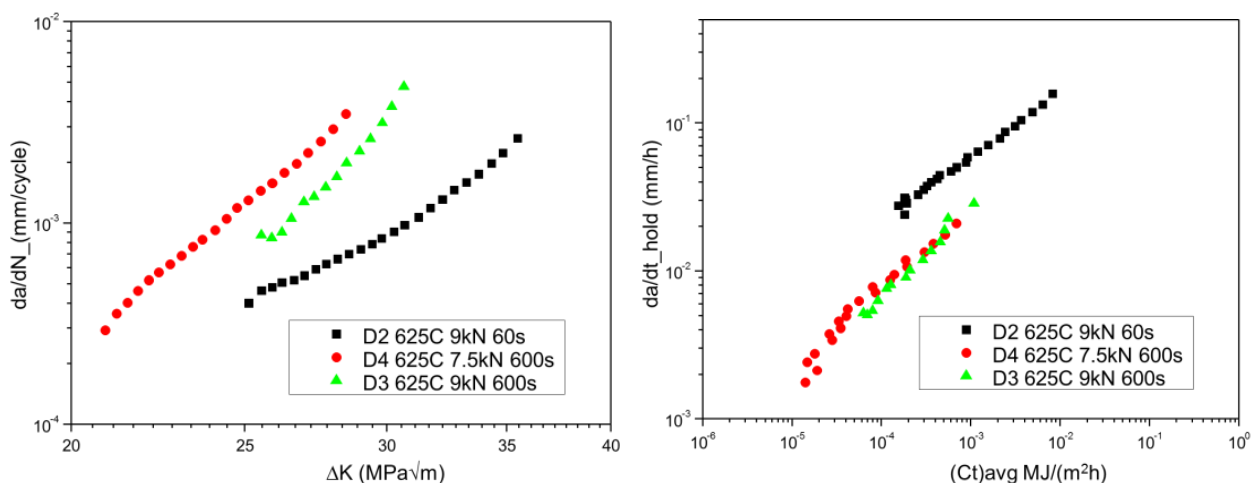


Figure 3.3.6.- Creep-fatigue crack growth data for T91-D material

Figure 3.3.6 shows the creep-fatigue crack growth data for T91-Demetra material. The effect of hold time and load is not so evident when ΔK is used as crack tip parameter, whereas when $(Ct)_{avg}$ is used, the effect of hold time is clear. The time rates of crack growth for the longer hold time tests of 600 seconds are lower than the crack growth rates from the 60 seconds hold time test; this difference is attributed to the time-dependent damage mechanisms at the crack tip. The load effect can be considered negligible.

Material T91-JRC

The results of the three tests are shown in the following tables.

Table 3.3.7: Creep-crack growth results. Material T91-JRC, specimen J1. 600°C, 9kN, hold 600 s

N	a	t	ΔVc	ΔVe	ΔK	da/dN_poly	Ct	da/dt_hold
Cycle	mm	h	mm	mm	MPa $\sqrt{m}$	mm/cycle	MJ/(m ² h)	mm/h
	20.65							
273	21.12	45.72	2.84E-04		25.22			
809	21.37	135.66	1.37E-04		25.56			
1381	21.62	231.64	8.59E-05	4.16E-06	25.90	5.18E-04	2.29E-05	3.11E-03
1962	21.86	329.13	2.06E-04	3.67E-06	26.25	4.45E-04	5.52E-05	2.67E-03
2528	22.11	424.09	2.13E-04	3.94E-06	26.61	4.65E-04	5.76E-05	2.79E-03
3121	22.37	523.59	2.03E-04	4.45E-06	27.00	5.10E-04	5.54E-05	3.06E-03
3588	22.63	601.95	2.90E-04	5.14E-06	27.40	5.71E-04	7.95E-05	3.43E-03
4030	22.87	676.12	3.18E-04	6.06E-06	27.78	6.55E-04	8.81E-05	3.93E-03
4403	23.12	738.70	2.00E-04	7.11E-06	28.18	7.48E-04	5.58E-05	4.49E-03
4729	23.38	793.41	3.45E-04	8.26E-06	28.60	8.44E-04	9.70E-05	5.06E-03
5003	23.62	839.38	3.40E-04	9.75E-06	29.01	9.68E-04	9.66E-05	5.81E-03
5254	23.87	881.49	5.07E-04	1.15E-05	29.43	1.11E-03	1.45E-04	6.68E-03
5490	24.11	921.09	6.11E-04	1.34E-05	29.87	1.25E-03	1.77E-04	7.51E-03
5665	24.37	950.46	7.64E-04	1.52E-05	30.33	1.38E-03	2.23E-04	8.29E-03
5828	24.62	977.81	9.96E-04	1.76E-05	30.78	1.56E-03	2.94E-04	9.33E-03
6004	24.86	1007.34	1.14E-03	2.10E-05	31.25	1.80E-03	3.39E-04	1.08E-02
6139	25.11	1029.99	1.34E-03	2.46E-05	31.73	2.04E-03	4.03E-04	1.22E-02
6254	25.37	1049.29	1.68E-03	2.80E-05	32.24	2.25E-03	5.11E-04	1.35E-02
6360	25.63	1067.07	2.35E-03	3.33E-05	32.77	2.59E-03	7.22E-04	1.55E-02
6454	25.87	1082.84	3.06E-03	4.00E-05	33.29	3.01E-03	9.51E-04	1.81E-02
6542	26.09	1097.61	3.87E-03	5.14E-05	33.76	3.76E-03	1.21E-03	2.26E-02
6606	26.36	1108.35	5.03E-03	6.90E-05	34.36	4.88E-03	1.60E-03	2.93E-02
6663	26.62	1117.91	7.62E-03		34.95			
6694	26.85	1123.12	1.02E-02		35.49			
6723	27.10	1127.98	1.29E-02		36.08			

Table 3.3.8: Creep-crack growth results. Material T91-JRC, specimen J2. 625°C, 9kN, hold 60 s

N	a	t	ΔVc	ΔVe	ΔK	da/dN_poly	Ct	da/dt_hold
Cycle	mm	h	mm	mm	MPa $\sqrt{m}$	mm/cycle	MJ/(m ² h)	mm/h
	20.65							
383	21.49	6.81	1.09E-04		25.77			
881	21.73	15.68	1.97E-04		26.11			
1441	21.99	25.64	2.13E-04	5.14E-06	26.49	6.12E-04	5.75E-04	3.67E-02
1972	22.24	35.09	1.13E-03	4.02E-06	26.86	4.65E-04	3.07E-03	2.79E-02
2533	22.48	45.09	2.98E-04	4.26E-06	27.23	4.79E-04	8.18E-04	2.87E-02
3084	22.72	54.89	2.75E-04	4.64E-06	27.60	5.08E-04	7.58E-04	3.05E-02
3540	22.98	63.01	1.43E-04	5.03E-06	28.01	5.35E-04	3.98E-04	3.21E-02
3973	23.22	70.72	7.40E-05	5.71E-06	28.40	5.91E-04	2.08E-04	3.55E-02
4408	23.48	78.47	4.08E-04	6.64E-06	28.83	6.66E-04	1.16E-03	4.00E-02
4796	23.70	85.38	1.93E-04	7.85E-06	29.21	7.67E-04	5.50E-04	4.60E-02
5117	23.99	91.09	6.33E-04	9.25E-06	29.70	8.74E-04	1.82E-03	5.24E-02
5397	24.24	96.08	5.82E-04	1.09E-05	30.15	1.00E-03	1.69E-03	6.00E-02
5626	24.48	100.16	8.11E-04	1.28E-05	30.60	1.14E-03	2.38E-03	6.82E-02
5860	24.73	104.32	4.35E-04	1.47E-05	31.06	1.27E-03	1.29E-03	7.61E-02
6020	24.96	107.17	1.40E-03	1.72E-05	31.50	1.45E-03	4.19E-03	8.68E-02
6187	25.22	110.15	1.65E-03	2.04E-05	32.01	1.66E-03	4.99E-03	9.95E-02
6341	25.43	112.89	1.94E-03		32.44			
6473	25.74	115.24	2.68E-03		33.07			
6588	25.99	117.29	3.55E-03		33.60			

Table 3.3.9: Creep-crack growth results. Material T91-JRC, specimen J3. 600°C, 7.5kN, hold 60 s

N	a	t	ΔV_c	ΔV_e	ΔK	da/dN_poly	Ct	da/dt_hold
Cycle	mm	h	mm	mm	MPa $\sqrt{m}$	mm/cycle	MJ/(m ² h)	mm/h
	20.65							
6	22.08	0.93	2.60E-04		22.27			
603	22.30	10.72	1.36E-06		22.54			
2549	22.38	45.35	1.65E-05	2.71E-06	22.65	3.67E-04	3.76E-05	2.20E-02
2962	22.83	52.70	2.15E-04	2.37E-06	23.23	3.05E-04	4.99E-04	1.83E-02
3861	23.09	68.71	-2.57E-06	2.76E-06	23.57	3.46E-04	-6.00E-06	2.08E-02
4468	23.30	79.52	6.81E-05	3.16E-06	23.86	3.87E-04	1.60E-04	2.32E-02
5306	23.59	94.44	-1.57E-04	3.02E-06	24.26	3.58E-04	-3.73E-04	2.15E-02
5940	23.83	105.73	9.07E-05	3.23E-06	24.61	3.71E-04	2.17E-04	2.23E-02
6624	24.07	117.91	-5.43E-05	3.51E-06	24.97	3.92E-04	-1.31E-04	2.35E-02
7244	24.33	128.95	5.45E-05	3.86E-06	25.36	4.18E-04	1.33E-04	2.51E-02
7925	24.59	141.08	1.13E-04	4.25E-06	25.76	4.46E-04	2.78E-04	2.68E-02
8389	24.83	149.33	5.90E-05	4.59E-06	26.14	4.67E-04	1.47E-04	2.80E-02
8878	25.08	158.04	-4.38E-05	4.96E-06	26.54	4.91E-04	-1.10E-04	2.94E-02
9402	25.32	167.37	5.20E-05	5.38E-06	26.94	5.17E-04	1.32E-04	3.10E-02
9865	25.56	175.61	7.62E-05	5.74E-06	27.36	5.35E-04	1.96E-04	3.21E-02
10324	25.82	183.78	1.35E-04	6.21E-06	27.81	5.59E-04	3.50E-04	3.35E-02
10789	26.08	192.07	8.99E-05	6.82E-06	28.27	5.94E-04	2.36E-04	3.57E-02
11207	26.34	199.51	1.03E-04	7.54E-06	28.75	6.35E-04	2.73E-04	3.81E-02
11605	26.58	206.60	2.03E-04	8.33E-06	29.21	6.80E-04	5.47E-04	4.08E-02
11969	26.84	213.08	1.39E-04	9.27E-06	29.72	7.31E-04	3.78E-04	4.39E-02
12278	27.08	218.58	2.16E-04	1.02E-05	30.21	7.80E-04	5.94E-04	4.68E-02
12613	27.34	224.55	1.60E-04	1.14E-05	30.74	8.43E-04	4.46E-04	5.06E-02
12893	27.58	229.53	3.62E-05	1.22E-05	31.26	8.68E-04	1.02E-04	5.21E-02
13144	27.81	234.00	4.19E-04	1.34E-05	31.77	9.26E-04	1.19E-03	5.55E-02
13417	28.08	238.87	5.69E-04	1.51E-05	32.38	1.01E-03	1.64E-03	6.03E-02
13686	28.30	243.66	6.29E-04	1.71E-05	32.88	1.10E-03	1.84E-03	6.59E-02
13901	28.58	247.49	5.17E-04	1.88E-05	33.57	1.16E-03	1.53E-03	6.97E-02
14097	28.83	250.98	4.82E-04	2.09E-05	34.17	1.25E-03	1.45E-03	7.49E-02
14297	29.09	254.54	9.29E-04	2.40E-05	34.84	1.38E-03	2.83E-03	8.26E-02
14489	29.32	257.96	1.31E-03	2.73E-05	35.44	1.51E-03	4.04E-03	9.07E-02
14649	29.57	260.81	1.35E-03	3.25E-05	36.15	1.73E-03	4.22E-03	1.04E-01
14788	29.83	263.28	1.53E-03	3.98E-05	36.87	2.04E-03	4.85E-03	1.22E-01
14925	30.09	265.72	2.20E-03	4.98E-05	37.63	2.45E-03	7.07E-03	1.47E-01
15009	30.32	267.22	2.76E-03	5.88E-05	38.32	2.79E-03	8.98E-03	1.67E-01
15096	30.58	268.77	3.34E-03	7.37E-05	39.12	3.35E-03	1.10E-02	2.01E-01

15174	30.83	270.16	4.78E-03	9.29E-05	39.96	4.05E-03	1.60E-02	2.43E-01
15231	31.05	271.17	5.54E-03	1.15E-04	40.67	4.83E-03	1.88E-02	2.90E-01
15283	31.33	272.10	6.37E-03	1.44E-04	41.66	5.77E-03	2.20E-02	3.46E-01
15325	31.58	272.85	9.30E-03	1.84E-04	42.54	7.07E-03	3.25E-02	4.24E-01
15361	31.84	273.49	1.24E-02	2.39E-04	43.50	8.81E-03	4.42E-02	5.29E-01
15386	32.04	273.93	1.59E-02	2.95E-04	44.27	1.05E-02	5.73E-02	6.29E-01
15412	32.34	274.40	1.86E-02	3.86E-04	45.48	1.30E-02	6.81E-02	7.79E-01
15428	32.58	274.68	2.01E-02	4.80E-04	46.49	1.55E-02	7.45E-02	9.29E-01
15447	32.84	275.02	2.56E-02	6.57E-04	47.61	2.02E-02	9.67E-02	1.21E+00
15458	33.07	275.21	3.16E-02		48.64			
15466	33.28	275.36	3.70E-02		49.64			
15475	33.58	275.52	4.40E-02		51.10			

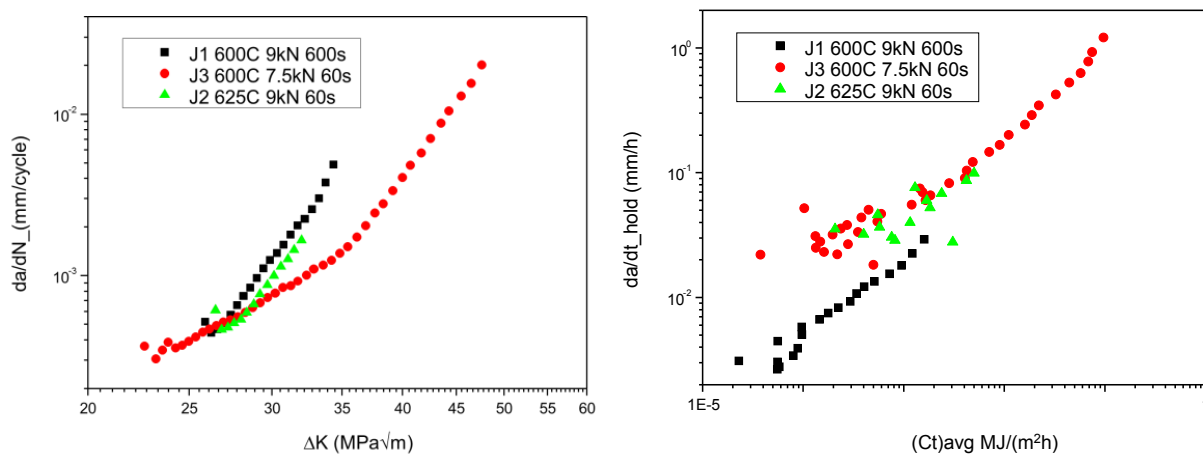


Figure 3.3.7. Creep-fatigue crack growth data for T91-RR material

As can be seen in Figure 3.3.7, specimen J3 shows a non-usual crack growth rate. For this specimens the signal of the extensometer was not as accurate as expected, due to an overheating during part of the test duration.

Material 316L

Only one specimen was tested of 316L. The results can be seen in Table 3.3.10

Table 3.3.10: Creep-crack growth results. Material 316L, specimen 3D1. 625°C, 9kN, hold 60 s

N	a	t	ΔV_c	ΔV_e	ΔK	da/dN_poly	Ct	da/dt_hold
Cycle	mm	h	mm	mm	MPa $\sqrt{m}$	mm/cycle	MJ/(m ² h)	mm/h
	20.65							
462	20.96	8.21	-3.68E-06		25.05			
949	21.23	16.87	5.05E-05		25.42			
1334	21.49	23.72	-2.08E-05	4.65E-06	25.78	5.85E-04	-5.54E-05	3.51E-02
1755	21.72	31.21	2.39E-05	4.96E-06	26.10	6.09E-04	6.41E-05	3.65E-02
2172	21.96	38.63	1.39E-04	5.62E-06	26.46	6.71E-04	3.76E-04	4.03E-02
2526	22.13	44.93	3.58E-05	6.29E-06	26.71	7.37E-04	9.72E-05	4.42E-02
2872	22.47	51.08	2.23E-04	7.17E-06	27.22	8.10E-04	6.12E-04	4.86E-02
3173	22.74	56.44	2.72E-04	7.78E-06	27.64	8.52E-04	7.52E-04	5.11E-02
3453	22.98	61.42	3.04E-04	8.20E-06	28.01	8.74E-04	8.47E-04	5.25E-02
3741	23.23	66.54	2.45E-04	8.27E-06	28.43	8.56E-04	6.88E-04	5.13E-02
4047	23.49	71.99	2.67E-04	8.85E-06	28.86	8.89E-04	7.57E-04	5.33E-02
4363	23.74	77.61	3.83E-04	1.00E-05	29.28	9.80E-04	1.09E-03	5.88E-02
4612	23.97	82.04	4.20E-04	1.11E-05	29.67	1.06E-03	1.21E-03	6.35E-02
4830	24.25	85.92	6.13E-04	1.26E-05	30.17	1.15E-03	1.78E-03	6.92E-02
5012	24.48	89.16	5.51E-04	1.43E-05	30.60	1.27E-03	1.62E-03	7.64E-02
5229	24.72	93.02	7.04E-04	1.63E-05	31.04	1.41E-03	2.09E-03	8.48E-02
5415	24.99	96.33	8.43E-04	1.80E-05	31.56	1.51E-03	2.53E-03	9.06E-02
5552	25.24	98.76	9.99E-04	1.94E-05	32.05	1.58E-03	3.03E-03	9.46E-02
5705	25.50	101.49	1.28E-03	2.17E-05	32.57	1.71E-03	3.93E-03	1.03E-01
5856	25.73	104.17	1.34E-03	2.43E-05	33.06	1.86E-03	4.14E-03	1.12E-01
6000	25.98	106.73	1.57E-03	2.75E-05	33.59	2.04E-03	4.91E-03	1.22E-01
6113	26.24	108.74	1.89E-03	3.14E-05	34.15	2.25E-03	5.99E-03	1.35E-01
6212	26.49	110.50	2.11E-03	3.49E-05	34.71	2.42E-03	6.76E-03	1.45E-01
6319	26.73	112.41	2.31E-03	3.96E-05	35.28	2.66E-03	7.46E-03	1.60E-01
6408	26.99	113.99	2.97E-03	4.37E-05	35.89	2.83E-03	9.73E-03	1.70E-01
6494	27.22	115.52	3.27E-03	4.80E-05	36.44	3.02E-03	1.08E-02	1.81E-01
6572	27.49	116.91	3.39E-03	5.29E-05	37.12	3.21E-03	1.14E-02	1.93E-01
6653	27.75	118.35	4.15E-03	5.66E-05	37.79	3.31E-03	1.41E-02	1.99E-01
6725	27.99	119.63	4.74E-03	5.84E-05	38.46	3.30E-03	1.63E-02	1.98E-01
6793	28.24	120.84	3.38E-03	6.04E-05	39.14	3.30E-03	1.18E-02	1.98E-01
6870	28.49	122.21	3.12E-03	6.39E-05	39.84	3.37E-03	1.10E-02	2.02E-01
6957	28.73	123.76	3.39E-03	6.92E-05	40.53	3.52E-03	1.21E-02	2.11E-01
7026	29.00	124.99	4.43E-03	7.67E-05	41.35	3.75E-03	1.60E-02	2.25E-01
7087	29.24	126.07	5.68E-03	8.46E-05	42.12	3.99E-03	2.08E-02	2.39E-01
7148	29.49	127.16	6.00E-03	9.60E-05	42.92	4.36E-03	2.23E-02	2.62E-01

7205	29.74	128.17	7.24E-03	1.08E-04	43.74	4.74E-03	2.73E-02	2.84E-01
7256	29.96	129.08	8.24E-03	1.24E-04	44.51	5.25E-03	3.14E-02	3.15E-01
7303	30.23	129.91	9.46E-03	1.49E-04	45.47	6.03E-03	3.66E-02	3.62E-01
7346	30.49	130.68	1.24E-02	1.73E-04	46.42	6.72E-03	4.88E-02	4.03E-01
7382	30.74	131.32	7.45E-03	2.05E-04	47.36	7.65E-03	2.97E-02	4.59E-01
7408	30.98	131.78	7.02E-04		48.29			
7445	31.25	132.44	-1.30E-03		49.39			
7463	31.47	132.76	1.12E-03		50.34			

In **Figure 3.3.8**, all the tests performed at 625°C, 9kN and 60 s of hold time are plotted together. As can be seen, the crack growth is similar for T91 and 316L material at this condition

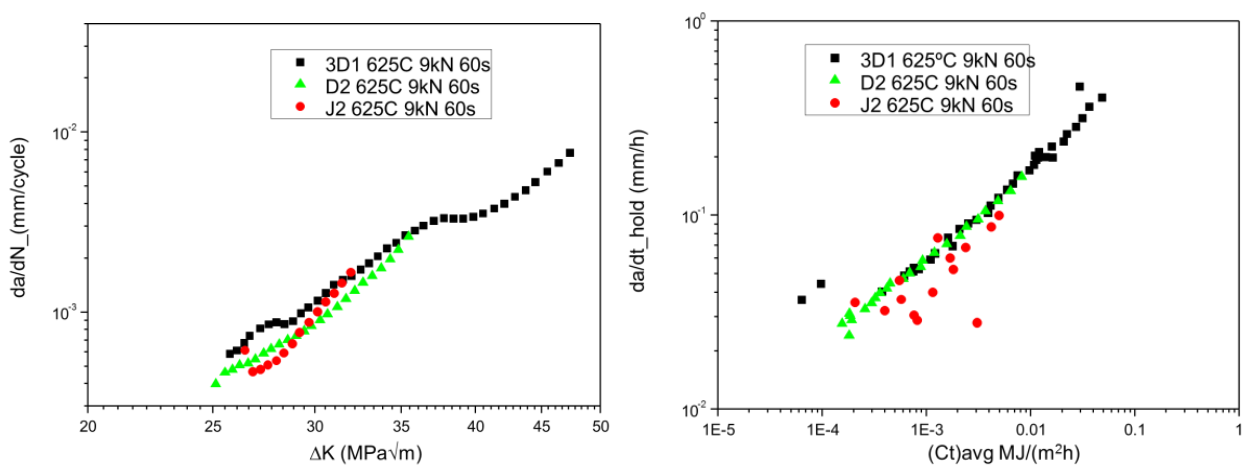


Figure 3.3.8. Creep-fatigue crack growth data for T91-F, T91-JRC and 316L material at 625°C, 9kN and hold time of 60s

3.4 CVR creep-fatigue tests

Investigation of creep-fatigue crack growth propagation was performed on steels P91 in the new laboratories of Research Centre Rez (CVR) in Pilsen (CZE). Creep-fatigue crack growth (CFCG) tests were performed on CT specimens which were made from a block of material (150x150x60 mm). 8 CT specimens were created, 4 pieces were sent to JRC and 3 were tested in CVR. The last was kept as a reserve. The dimensions are shown in **Fig. 3.4.1**. The arrow on the block indicates the rolling direction. The chemical composition is defined in **Table 3.4.1**. Chemical composition was also analysed by glow discharge spectroscopy (see **Table 3.4.2**).

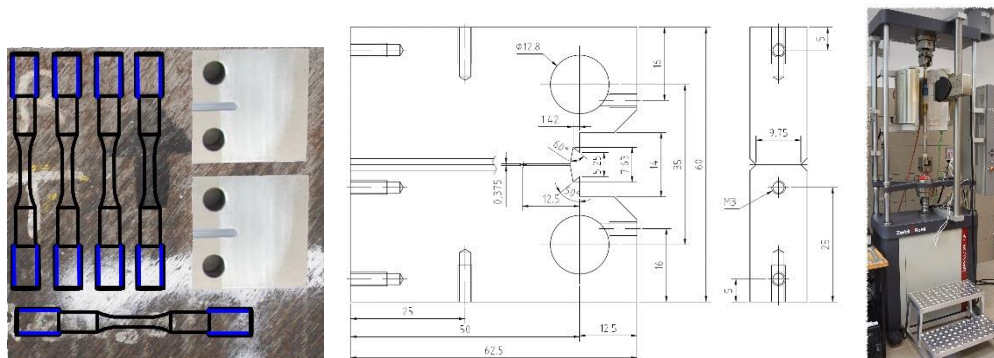


Fig. 3.4.1 Cutting plan of block P91 (left), the shape and dimensions of specimen (centre), and testing machine Kappa SS-CF (right).

Table 3.4.1 Chemical composition of the test material

Elements	C	Mn	Si	Ni	Cr	Mo	Cu	Al	S	P
unit [%]	0.12	0.41	0.24	0.10	8.32	1.02	0.05	0.006	0.001	0.009
Elements	V	Nb	N	As	Sb	Ti	B	W	Zr	O
unit [%]	0.235	0.084	0.041	0.005	0.001	0.002	0.0009	0.001	0.001	15 ppm

Table 3.4.2 Chemical composition analysed by glow discharge spectroscopy.

		C	Si	Mn	S	P	Cu	Ni	Cr	Mo	Ti	Nb
P91	% wt	0.097	0.202	0.347	< 0.001	0.007	0.062	< 0.1	7.6	0.885	0.014	0.061
	SD	0.002	0.003	0.003	-----	0.001	0.0005	-----	0.1	0.007	0.0005	0.001

The specimens were pre-cracked in air to an initial crack length of approximately 20 mm (notch included 12.5 mm). Fatigue pre-cracking was conducted at a load ratio R of 0.1. The resonant testing machine TESTRONIC 250 made by the Swiss firm RUMUL was used for pre-cracking. It is a dynamic testing machine, which works in full resonance for loads up to 250 kN. The maximum oscillating stroke extension available is up to 4 mm. The resonant operating frequency ranges from 40 Hz up to 250 Hz provided by the oscillating masses and also by the stiffness of the specimen. The frequency was around 90 Hz. The force peak to peak was determined from requirement for $\Delta K = 20$ and ranged from 11.34 kN to 7.65 kN with the decrease step of 1 kN as the crack size grew. The final crack size from pre-cracking was around 7.5 mm for all 3 samples, as seen in **Fig. 3.4.2**. The number of cycles was from 170 up to 220 thousand. Side grooves were creating after pre-cracking.

**Fig. 3.4.2** The value of crack size after pre-cracking is 7.5 mm.

Further testing was performed on electromechanical creep testing machine Kappa SS-CF which offers a wide range of applications especially creep-fatigue tests through-zero (see **Fig. 3.4.1**). Load capacity is up to 50 kN and speed range is 1 $\mu\text{m/h}$ to 100 mm/min. The machine was equipped with a 3-zone furnace to 1200°C and non-contact high resolution video extensometer which was in this case replaced by a contact COD extensometer. The ratio of load asymmetry was 0.1 for all three tests. The parameters and results are summarized in **Table 3.4.3**. Time ramped up (t_r) was same as time down (t_d) and it was equal to 2 seconds.

Table 3.4.3 Parameters and results of CFCG tests.

Specimen	Temp [°C]	Hold time [s]	Pmax(kN)/R	B [mm]	B_N [mm]	W [mm]	a0 [mm]	a [mm]	Δa [mm]	N [cyc]
VZ01	600	600	9/0.1	12.5	9.75	50	20.6	21.9	1.3	3027
VZ02	600	60	9/0.1	12.5	9.75	50	21.0	22.1	1.1	3032
VZ03	600	0	9/0.1	12.5	9.75	50	20.6	21.6	1.0	3169

The end of the test was chosen after 3000 cycles due to time pressures and on the basis of other results from Ciemat. The initial crack size (a_0) and the final crack size (a) was determined from the fracture surfaces after final break (see **Fig. 3.4.3**). The break was performed in air by fatigue rupture on the same resonant machine as pre-cracking.

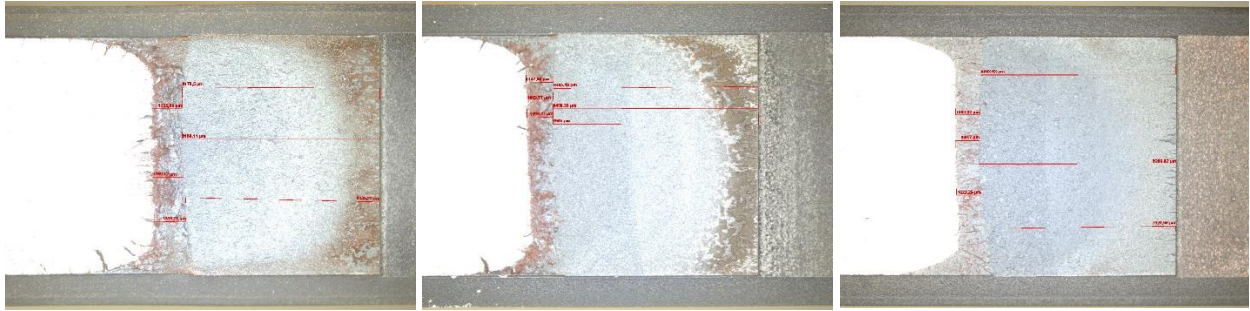


Fig. 3.4.3 Fracture surface VZ01 with hold time 600s, VZ02 with hold time 60s and VZ03 with hold time 0s succession order.

The tests outputs are the values of load line displacement (VLL). Unfortunately the data from DCPD is missing due to a software error. The differences between the tests with different hold times are shown in **Fig. 3.4.4**.

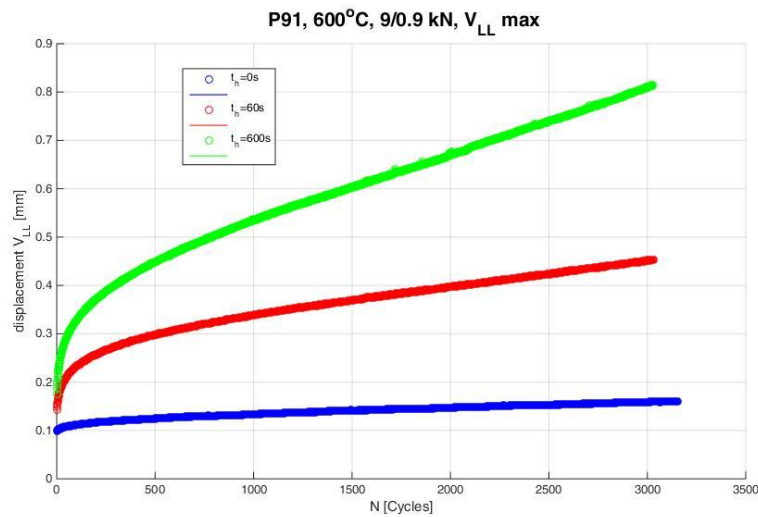


Fig. 3.4.4 Comparison of load line displacement data for different hold times.

It is possible to compute the length of crack size (a) from values of VLL. According to ASTM standard E647 the normalized crack size is set as a function of plane stress elastic compliance for C(T) specimen in the form:

$$a = (C_0 + C_1 u_x + C_2 u_x^2 + C_3 u_x^3 + C_4 u_x^4 + C_5 u_x^5) \cdot W$$

where W is specimen length,

$$u_x = \left\{ \left[\frac{EVB}{P} \right]^{\frac{1}{2}} + 1 \right\}^{-1}$$

and E is elastic modulus of the sample (at 600 °C in this case), B is specimen thickness, P is force and V is COD evaluation.

Determination of stress-intensity factor range (ΔK):

$$\Delta K = K_{\max} - K_{\min}$$

where K – stress intensity factor for CT specimen subjected to Mode I loading is calculated by the following equation:

$$K = \frac{P}{(BB_N)^{1/2} W^{1/2}} f(a/W) [\text{Nm}^{-3/2}].$$

The results of computations for determination of crack size are shown in **Fig 3.4.5**. The value of 130 GPa was substituted as elastic modulus. These results do not correspond with real results from the fracture surface. For example, for hold time 600s the crack starts at 22.1 mm and ends at 35.3 mm instead of the correct values of 20.6 to 21.9 mm. For hold time 60s the crack starts at 19.8 mm and ends at 31.1 mm instead of the correct values of 21.0 to 22.1 mm. And for hold time 0s the crack starts at 16.2 mm and ends at 21.3014 mm instead of the correct values of 20.6 to 21.6 mm. This is a topic for further consultation. On the grounds of this and finding out that tests are at the beginning of the crack growth curve, the next computations for stress-intensity factor range are not presented.

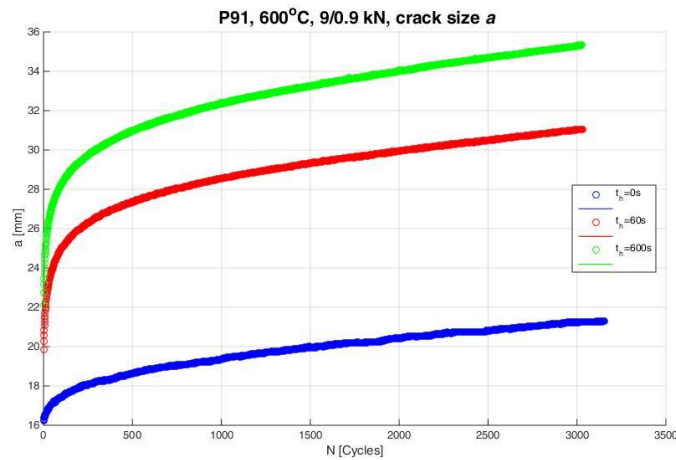


Fig. 3.4.5 Comparison of the results of computations for determination of crack size.

4 JRC Thermal Fatigue Tests

4.1 Introduction

Structural components in fast nuclear reactors are primarily exposed to thermo-mechanical loads. Typical thermal loadings on the component can be described by a spectrum in the case of a mixing hot and cold fluids or a given frequency for instance in the above core structure of a sodium fast reactor where exposed to frequency cold sodium jets, moving free surface or start-ups and shutdown with a very low frequency. The resulting through-wall temperature distribution is characterized by cyclic temperature gradients with the largest temperature amplitudes at the surface. Since the material cannot expand freely due to of the component the temperature gradients give rise to thermal stress gradients where a cold shock gives tensile stresses and a hot shock compressive stresses. Thermal stresses are self-equilibrating, i.e. there is no resultant force when the stresses are integrated over the wall thickness. As a consequence stresses at the outer surface respond to thermal loads in the inner surface well before the temperature change from heat conduction. Cracking is often initiated early due to locally high surface stresses but the actual crack propagation is much slower and cracks may also completely arrest at a certain depth as the stress range gets smaller way from the heated surface. The total fatigue life of the component is therefore mainly controlled by crack propagation. In addition to the thermal loads there may also be primary loads from for instance internal pressure which gives an increase of the mean stress which affects crack growth. For components that operate in the creep regime crack propagation may also have a creep component, thus we may be faced with thermal creep-fatigue. To assess this JRC performed thermal fatigue test of pipe components of 316L and P91 steels where the initiation and growth of cracks was monitored by non-destructive inspection methods. These tests were then modelled by a combined thermal and mechanical analysis to predict the crack propagation and life of the components. The work has been published in more detail in [12], so the information in this Deliverable is summary of that paper.

4.2 Experimental Programme

The thermal fatigue test rig and the pipe geometry are shown in **Figure 4.1**. In this study the central part of the pipe's outer surface is heated continuously by induction to a temperature of 550°C using an external heating coil. A thermal load cycle is induced by pumping room temperature water through the pipe for around 10 seconds. After the water flow has stopped, air is blown through the pipe to remove remaining water from the surface and the pipe heats up again through conduction from the hot outer surface for about 45 seconds when the next water flushing is started. The full thermal cycle takes around one minute. The pipe is held in a lever arm test machine providing an axial load, but no restraint on axial displacement. The thermal loads induce secondary self-equilibrating cyclic stresses whereas the axial load gives a primary stress. A test specimen equipped with five thermocouples across the wall thickness was used to measure the temperature profiles for P91 and 316L pipes respectively. The initial surface cracking was identified by replica methods. Two different non-destructive methods were used to measure crack growth: time-of-flight-diffraction (TOFD) [14], as a relatively quick method to determine crack depth and X-ray computed tomography (XCT) for a three-dimensional mapping of the damage [15].

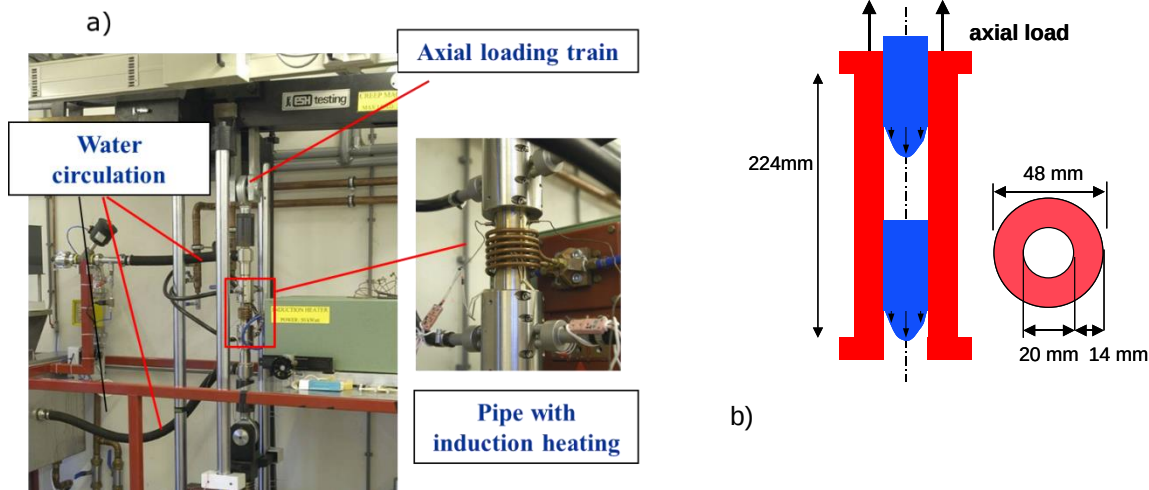


Table 4.1 summarizes the test matrix and measured crack depths.

- F is the imposed axial load: 0, 50, 100 and 150 kN, which correspond to a nominal axial stresses of 0, 33, 66 and 100 MPa respectively.
- N_i is number of cycles after which crack initiation was detected by replica. For the first tests (TFR1-TFR5) the first check was performed after 10,000 cycles following the same procedure adopted for previous tests, where the maximum temperature was 350°C. With 550°C surface cracking occurred well before 10,000 cycles, hence the "<" sign, and the first replica test were subsequently taken already after 1,000 cycles.
- N_{test} is the number of conducted thermal cycles.
- a_{max} refers to the maximum crack depth measured at N_{test} for circumferential and/or axial cracks a (defined in **Figure 4.2**). Failure means that a through-wall crack has formed. Good agreement was observed between crack depths from TOFD and CT results.

Test	Mat.	F (kN)	N_i (kcyc)	N_{test} (kcyc)	a_{max} (mm)/failure	
					Circ	Axial
TFR1	316L	0	<10	46	6.3	failure
TFR3	316L	0	<10	90	12.8	failure
TFR2	316L	50	<10	58	failure	7.2
TFR5	316L	50	<10	30	10.7	9.3
TFR15	316L	100	<1	5	8	0
TFR16	316L	100	<1	3	5.5	Very small
TFR12	316L	150	<1	9	failure	3.5
TFR14	316L	150	<1	6	7	Very small
AB04	P91	150	6-8	20	5	Very small
AB05	P91	150	6-8	23	7	Very small

Table 4.1 Summary of Thermal fatigue test

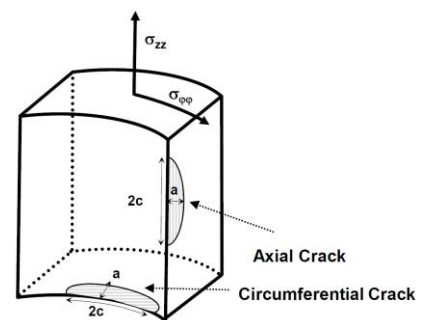


Figure 4.2. Definition of circumferential and axial cracks and axial (σ_{zz}) and hoop stress ($\sigma_{\phi\phi}$)

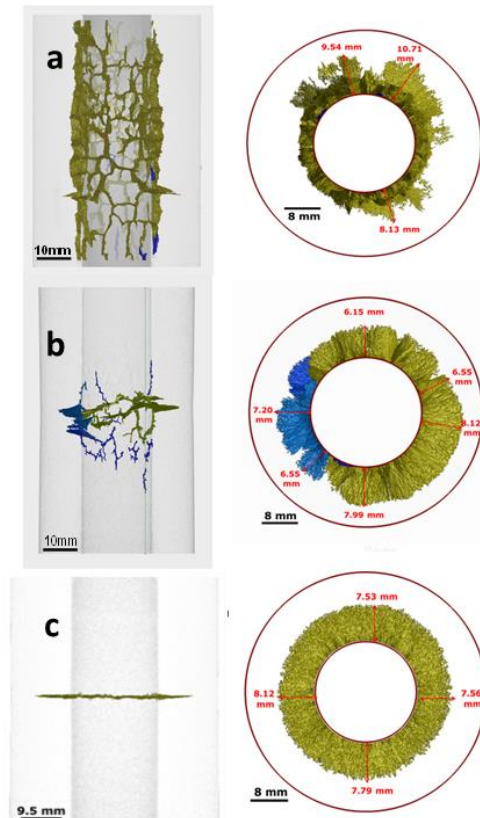


Figure 4.3 Mapping of crack configuration: a) TFR 5, $F = 50\text{ kN}$, 30000 cycles; b) TFR 12, $F = 150\text{ kN}$, 7000 cycles; c) TFR15, $F = 100\text{ kN}$, 5000 cycles

Cracking is initiated at the inner surface as a network of shallow cracks followed by growth of some cracks. It follows from **Table 4.1** that the number of cycles needed to reach a specific crack depth is generally reduced by increasing the axial load. The axial load promotes the growth of circumferential cracks and reduces the number of deep cracks. This is clearly seen from the XCT mappings of TFR5, TFR12 and TFR 15 in **Figure 4.3**. For TFR 5 with $F = 50\text{ kN}$, a very large number of interlinked axial cracks can be observed together with a small number of circumferential cracks. For TFR12 the two circumferential cracks are dominant whereas for TFR15 one, an almost circumferential planar crack, can be seen. It is surprising that the trend with a dominating circumferential crack is stronger for TFR 15 with $F = 100\text{ kN}$ than for TFR12 with $F = 150\text{ kN}$.

By comparing TFR12 and 16 with AB04 and AB05 in **Table 4.1** we see that initiation times are longer for P91 specimen than for 316L.

4.3 Modelling

The thermal fatigue life assessment is done in three consecutive steps. In the first step the temperature gradients are computed from a thermal analysis. In the second step the stresses and strains resulting from the thermal gradients and axial loads are computed. Both the thermal and stress analyses are done by finite element analyses of an uncracked component using the commercial FE-code ABAQUS. In the third step the crack initiation and propagation are calculated from the stress and strain fields in step 2 by post-processing procedures. The temperature dependent materials properties for 316L and P91 steels were taken from RCC-MRx material data sheets [12] if not stated otherwise. The details of each of these steps are given below.

4.3.1 Thermal analysis

The temperature distribution from the cyclic thermal loadings in the tests was calculated by the ABAQUS FE axisymmetric model. The finite element mesh at the inner surface was very dense to capture the strong thermal gradients. The time dependent temperature distribution is then governed by the heat transfer coefficients and the thermal conductivity of the two steels. The boundary conditions are illustrated in **Figure 4.4**. The segment of the outer surface of the pipe with the coil had a prescribed temperature of 550°C to represent the induction heating. During the cyclic thermal loading a temperature of 550°C was prescribed on the outer surface covered by the induction coil. For the cooling stage water with a temperature of 100°C and a constant heat transfer coefficient of $25\text{ kW/m}^2\text{K}$ was adopted for both steels. **Figure 4.5** shows the temperature distributions during the cooling and heating part of a stabilized cycle for 316L. The strongest temperature gradients occur at the beginning of the cooling whereas the lowest temperatures are attained at the end of the cooling.

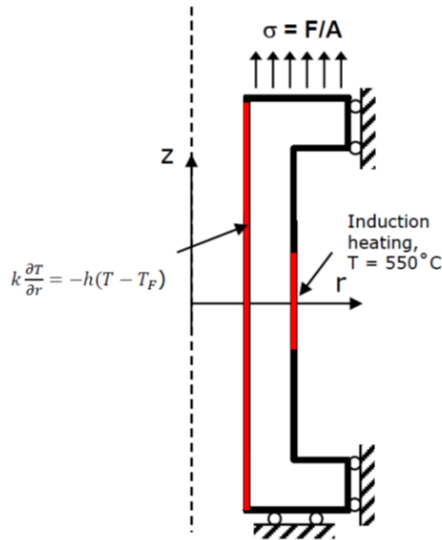


Figure 4.4 Illustration of applied boundary conditions during thermal and stress finite element analysis

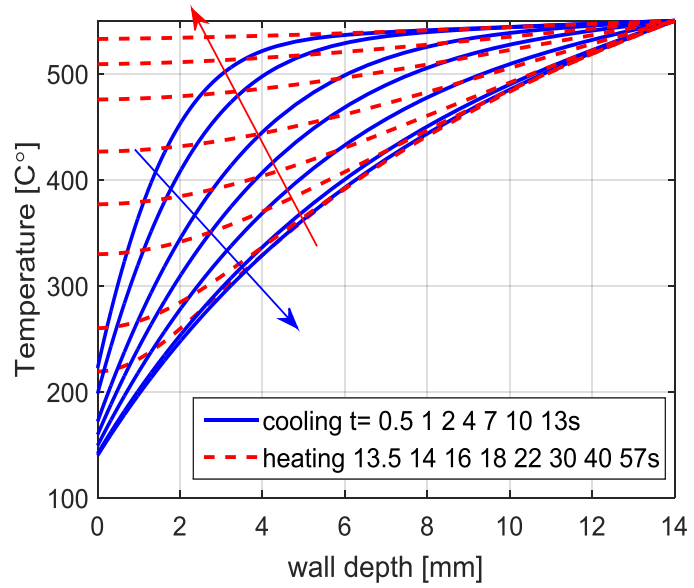


Figure 4.5 Computed Temperature distributions for the 316L pipe during cooling and heating.

4.3.2 Stress Analysis

Figure 4.4 also shows the used mechanical boundary conditions. The lower part is constrained axially; the upper part has a prescribed axial stress corresponding to the axial load. The lower and upper thicker parts of the pipe are constrained in the radial direction and all other parts are traction free.

The temperature dependent Young's modulus and Poisson's ratio were taken directly from the RCC-MRx data sheets for the two steels [10]. The thermal loadings induce low-cycle fatigue, which requires a cyclic plasticity constitutive model. For this we used the Chaboche model [16,17] implemented into the ABAQUS code. The cyclic plastic stress-strain curve can then be expressed as:

$$\frac{\Delta\sigma}{2} = \sum_{i=1}^M \frac{C_i}{\gamma_i} \tanh\left(\gamma_i \frac{\Delta\epsilon_i^p}{2}\right) + \sigma_y + R_0(1 - e^{-bp}) \quad (4.1)$$

Here $\Delta\sigma/2$ and $\Delta\epsilon_i^p/2$ are the stress and plastic strain amplitudes respectively, σ_y is the yield stress, $\sum_{i=1}^M \frac{C_i}{\gamma_i} \tanh(\gamma_i \Delta\epsilon_i^p)$ is the nonlinear kinematic hardening and $R_0(1 - e^{-bp})$ is the isotropic cyclic softening/hardening, where p is the accumulated cyclic plastic strain, R_0 is the asymptotic value for stabilized cycles and b controls the speed of stabilization, and M is the number of terms used to describe nonlinear kinematic hardening. The material constants C_i , γ_i , σ_y , R_0 and b were calibrated to cyclic stress-strain data.

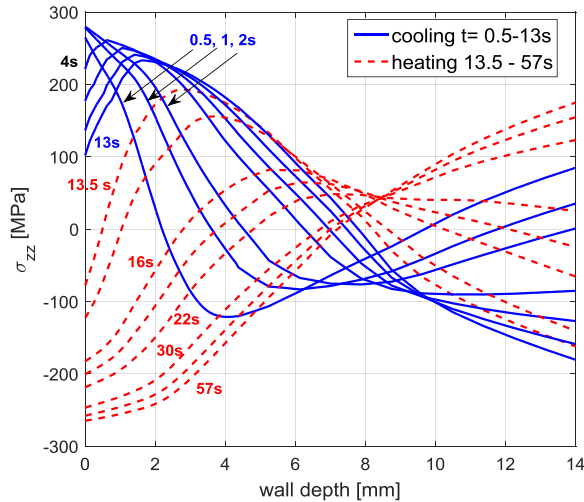


Figure 4.6 Computed axial stress through-wall distribution for 316L and no axial load. The thermal load cycle is 57s including cooling for 13 seconds.

Figure 4.6 shows the through-wall axial stress distributions for 316L, in the case with no axial load, from the onset of the down-shock until the end of the thermal cycle. At the end of the heating ($t = 57$ s), the inner surface is in compression and the outer surface in tension. The down-shock results in a very strong tensile stress increase close to the inner surface ($t = 0.5$ s) and stress reduction at the outer surface, but at a slower rate. The depth with tensile stress then spreads from the inner surface as the cooling proceeds, and when the cooling stops (13 s) and the temperature increases, the inner surface goes into compression and the outer surface into tension. Although the temperature gradient at the onset of cooling is confined to the inner surface, the entire stress distribution is affected since the secondary thermal stresses are self-equilibrating.

4.3.3 Fracture Analysis

4.3.3.1 Crack initiation

The prediction of crack initiation and propagation of short cracks was based on strain controlled low-cycle fatigue curves. **Figure 4.7** shows the computed total mechanical strain ($\epsilon_{mech} = \epsilon_{total} - \epsilon_{thermal}$) ranges versus crack depth for the axial and hoop strain component for 316L and P91 for the case $F = 0$ kN. The mean strain range between 0 and 1 mm depth is about 0.6% for 316L and 0.35% for P91. The near inner surface strain range for the hoop and axial component is very similar for both materials. The fatigue curves are temperature dependent and the fatigue life decreases with increasing temperature. From the RCC-MRx Design curves, the number of cycles to failure (typically defined as a load drop of 50% from cracks) for a strain range of 0.35% for P91 at 550°C, 500°C, 450°C and 20°C, are 3900, 5700, 8000 and 18000 respectively. For 316L the number of cycles to failure at a strain range of 0.60% at 550°C, 500°C, 450°C and 20°C, are 415, 612, 860 and 1900 respectively.

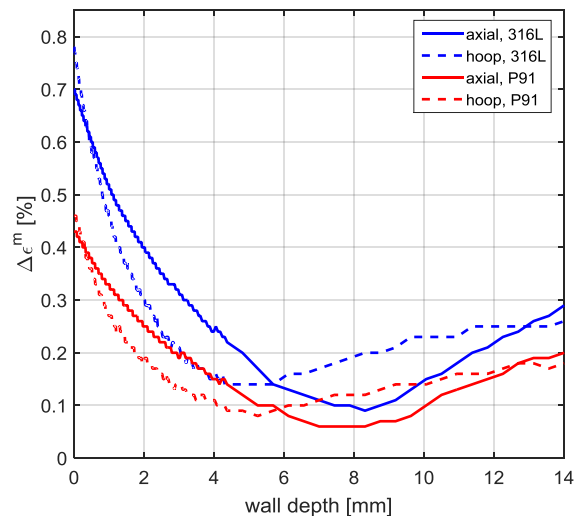


Figure 4.7 Computed distribution of the total mechanical strain range for axial and hoop components, 316L and P91 with $F = 0$.

4.3.3.2 Crack Propagation

A finite element simulation of interacting fatigue cracks with complex shapes and where cracks are modelled explicitly is impossible in practice. Therefore we adopted a number of simplifications for the crack propagation analysis. First of all we assumed a single circumferential or axial crack. We also assumed that the crack is semi-elliptical as shown in **Figure 4.8** or axi-symmetric when the length of a circumferential crack ($2c$) reaches the inner surface perimeter.

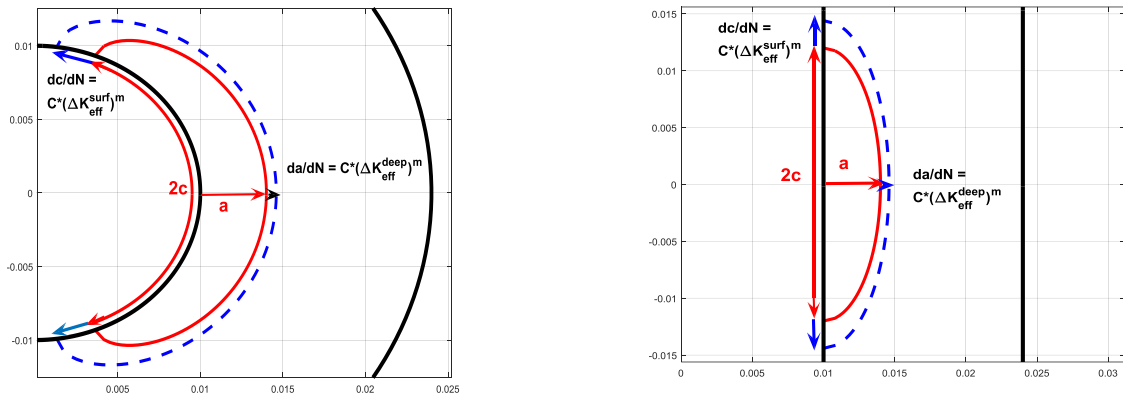


Figure 4.8. Model used for semi-elliptical circumferential (left) and axial crack (right).

The elastic stress intensity factor were computed for set of semi-elliptical or axi-symmetric cracks in a cylinder can be computed using tabulated K-solutions from RCC-MRx and the computed stress distribution. As the computed stresses were in the plastic range, plasticity corrections were applied. The growth of the fatigue crack was calculated from Paris Law,

$$(da/dN)_f = C_P (\Delta K_{eff})^{m_P} \quad (4.2a)$$

$$\Delta K_{eff} = q(K_{max} - K_{min}) \quad (4.2b)$$

$$q = \begin{cases} \frac{1}{1-R/2}; R > 0 \\ \frac{1-R/2}{1-R}; R < 0 \end{cases} \quad (4.2c)$$

The stress intensity factors were computed at the deepest point as well as at the outer edge. The values of Paris law parameters for 316L and P91 were also taken from RCC-MRx. **Figure 4.9** shows examples of computed stress intensity factors at the deepest point of the crack.

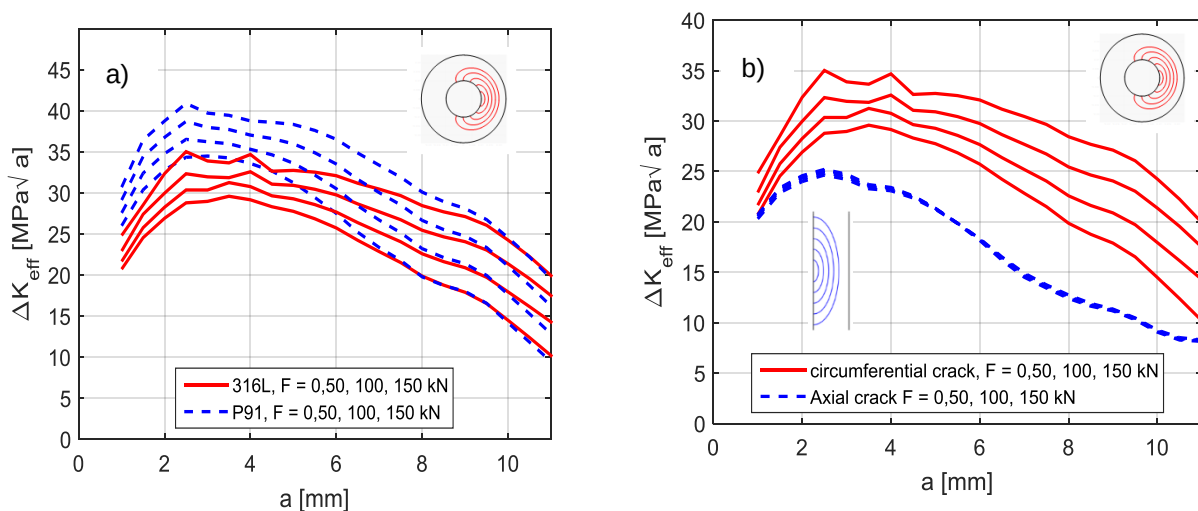


Figure 4.9 Computed maximum stress intensity factor range versus crack depth for $c/a = 2$ a) circumferential crack 316L and P91 b) circumferential and axial crack

The $C(t)$ were also computed from the stress intensity factor as function of crack depth and time using Eq. (2.3) to allow assessment of creep crack propagation and with the creep crack propagation governed by:

$$\left(\frac{da}{dN}\right)_c = \int_0^T \frac{K(a,t)^2(1-\nu^2)}{(n+1)Et} dt, \quad (4.3)$$

With the given material properties it turned out that the creep part was negligible compared to the fatigue contribution. The main reason is that high stress intensity factors are high for a very short time so the integrated effect is low over a cycle. Moreover at 550°C the creep rate is significantly lower than at 600°C and 625°C as in the standards tests in section 3.

For the crack propagation, we computed the stress intensity factors at the deepest point and the surface point for a single semi-elliptic or axis-symmetric crack. The crack propagation was calculated by cycle-by-cycle integration for the deepest and surface points,

$$\left.\begin{aligned} (da/dN)_f &= C_P(\Delta K_{eff}^{deep})^{m_p}, & a_{i+1} &= a_i + \frac{da}{dN}, \\ (dc/dN)_f &= C_P(\Delta K_{eff}^{surf})^{m_p}, & c_{i+1} &= ac_i + \frac{dc}{dN}, \end{aligned}\right\} \quad (4.4)$$

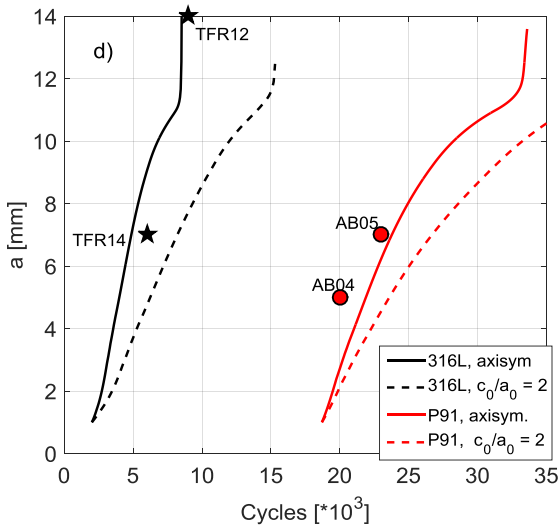
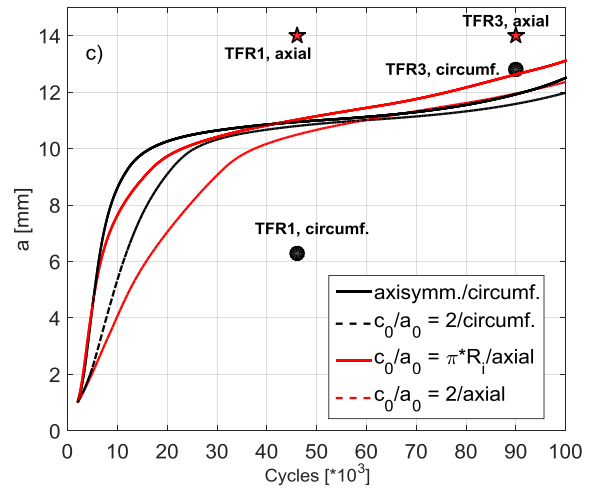
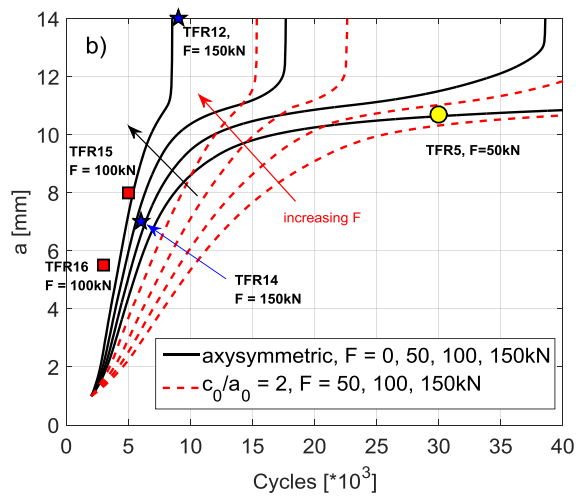


Figure 4.10 Computed crack depth versus number of cycles together with selected experimental data a) 316L circumferential crack with $c_0/a_0 = 2$ and axisymmetric b) 316L circumferential and axial crack $c_0/a_0 = 2$ and πR_i c) 316L and P91 circumferential crack $c_0/a_0 = 2$ and axisymmetric.

This approach simulates in a natural way the change of the crack shape. When the length along the inner surface for axial crack reaches the length of the pipe's inner perimeter (i.e. $2c=2\pi R_i = 62.83\text{mm}$), R_i the inner radius of the pipe, we switched from the Handbook solution of a semi-elliptic crack to an axi-symmetric crack. The depth of the initial crack from which we start the crack propagation, a_0 , is specified to be 1 mm. The crack propagation rate depends also on the specified initial crack aspect ratio. The total fatigue life is then computed by assuming a starter crack of a depth of 1 mm after the number of cycles predicted from the fatigue curve followed by crack propagation. **Figure 4.10** shows computed crack depths versus number of cycles for some different cases together with some experimental data. In **Figure 4.10a** the predictions are compared for the axi-symmetric and semi-elliptical starter crack ($c_0/a_0 = 2$). The initial defect shape has a large influence. A very good agreement with experimental data are achieved when an axisymmetric starter crack is used for the cases with a larger axial load, which is in agreement with the observed crack shape in **Figure 4.3** for TFR12, TFR 15 and TFR16. TFR5 agrees better with the $c_0/a_0 = 2$ which better correspond to the shape for TFR5 in **Figure 4.3**. For the cases TFR 1 and TFR3 failure was by the axial crack but axial and circumferential crack were of similar depths which is also in agreement with **Figure 4.10b** where more or less the same crack growth rate is predicted for axial and circumferential cracks. Finally **Figure 4.10c** show that the longer life for P91 is well captured. The difference lies however mainly in the superior resistance to crack initiation.

4.4 Discussion

These experiments simulated typical development of crack initiation followed by crack propagation for components subjected to cyclic thermal loads. The crack evolution is quite complex to simulate due to strongly varying stress and strain distributions, complex crack configuration and visco-plastic properties. Nevertheless it was possible to predict the fatigue for both 316L and P91 under the different loading conditions. An important observation is that crack propagation is controlled by fatigue; the creep effects are very small at 550C and the main reason for this is that peak stresses only occur for short times whereby creep crack growth is limited.

5 KAERI Tests Creep crack Growth, Tensile and Fracture Toughness P91 (base, weld and aged)

Test matrix of KAERI (Planned vs. Achieved)

Test item	Planned (originally)	Achieved	Remarks
[1] Creep Crack Growth Rate tests for Gr.91 Steel	600°C BM 600°C WM 600°C HAZ	600°C BM 600°C WM 600°C HAZ 550°C BM 550°C WM 550°C HAZ	✧ All planned CCG tests were completed. ✧ Extra tests conducted : for BM, WM & HAZ @550°C
[2] High temperature material strength tests for Gr.91 steel	500, 550, 600°C (virgin material) @ 1 strain rate 500, 550, 600°C (aged material) @ 1 strain rate	RT, 200, 250, 300, 350, 370, 400, 450, 500, 550, 600, 650°C for virgin and service exposed spec. @ three different strain rates - $6.67 \times 10^{-3}/s$, - $6.67 \times 10^{-4}/s$, - $6.67 \times 10^{-5}/s$.	✧ All planned tests were completed. ✧ Extra tests conducted for Gr.91 virgin & service exposed mat (73,716 h @USC plant material) conducted for RT, 200, 250, 300, 350, 370, 400, 450 & 650°C.) @ three different strain rates.
[3] J-R tests for Gr.91 steel	500, 550, 600°C	RT, 300, 370, 400, 425, 500, 550, 600°C for virgin and service exposed specimens	✧ All planned tests were completed. ✧ Extra tests for virgin & service exposed Gr.91 steel were conducted for RT, 300, 370, 400 and 425°C. ✧ Since J-R showed min value near 400°C, J-R tests for 370°C & 425°C were conducted.

5.1 CCG test results of welded Gr.91 specimens at 550°C and 600°C

The main components of Modified 9Cr-1Mo steel (ASME grade 9Cr-1Mo, hereafter Gr.91) in Generation-IV nuclear reactor systems are designed to last up to 60 years at elevated temperatures and will be generally subjected to non-uniform stress and temperature distribution during a long-time service. These conditions may generate localized creep damage, crack initiation and crack propagation which may lead high temperature fracture. A significant portion of their lives will be spent during crack propagation. It is, therefore, necessary to evaluate creep crack growth (CCG) behaviour during creep loading for design and safety assessment of the components, especially for Type IV cracking at heat affected zone (HAZ) in welded joint of high-Cr ferrite and martensite (FM) steels. To prevent such unexpected failures, an accurate residual life assessment for the welded joints is required.

Generally, welded joints are considered as the composite structure of different materials consists of base metal (BM), weld metal (WM), and HAZ. Since the inhomogeneity among those materials affects the state of stress field or strain rate field near the weld fusion line and near the BM/HAZ interface, it is not simple to estimate the crack propagation behavior at the welded joint. Also, the difference in creep properties for BM, WM, and HAZ may cause a difficulty in the assessment of crack behavior. Therefore, it is necessary to clarify the creep crack growth laws through experimental CCG tests for the welded joint of Gr.91 steel.

This report is to comparatively investigate the creep crack growth rates (CCGR) for BM, WM, and HAZ at 550°C and 600°C in the weldment of Gr.91, which is prepared using a shielded metal arc weld (SMAW) process. Creep crack growth laws at 550°C and 600°C for the BM, WM, and HAZ are newly constructed using a fracture parameter of C^* . The tested CCGR lines are compared with RCC-MRx code.

5.1.1 Basic material tests associated with CCG tests

The Gr.91 steel used in this report was a commercial type hot rolled plate with thickness of 32mm. Heat treatment conditions were normalized and tempered (N+T) at 1050°C /1min per mm and 770°C/3min

per mm. The groove shape of the welding of two plates was designed as a single V-groove with 60 degrees. Welded blocks were prepared by using the shielded metal arc welding (SMAW) process. The filler metal, CM-9Cb (brand name), was manufactured by Kobe steel as AWS Class, E9016-G (3.2-4.0 mm). The post weld heat treatment was maintained for 255 min at 750°C.

To obtain the material properties for the BM, WM, and HAZ regions, a series of tension and creep tests were performed at isothermal temperature of 550°C and 600°C. The HAZ specimens were cut out toward the transverse direction against the welding direction (longitudinal direction) in the welded block. The tension specimens had a rectangular cross section 6.25 mm in width with of 2 mm in thickness with a 25 mm gauge length. The strain rate was 1×10^{-4} /s at 550°C and 600°C. The creep specimens had a cylindrical shape with a 30 mm gauge length and 6 mm diameter. The HAZ location in the HAZ specimen was located at the centre of gauge length. Using constant-load creep machines of a dead-weight type with lever ratio of 20:1, the creep tests were carried out with different stress levels at the two temperatures of 550°C and 600°C. Creep strain data with elapsed times were taken automatically by a PC through a high precision LVDT. The steady state creep rate was measured from the secondary creep region of experimental creep curves. The experimental procedures for the creep tests followed the recommendations of the ASTM standard E139. From these tension and creep tests, the material constants of D, m, A, and n were obtained for the BM, WM, and HAZ samples at 550°C and 600°C.

The CCG tests for the BM, WM, and HAZ samples were conducted at a constant load with different applied load levels at 550°C and 600°C. Compact tension (CT) specimens had a width (W) of 25.4 mm, a thickness (B) of 12.7 mm, and side grooves of a 10% depth. The initial crack ratio (a/W) was about 0.5, and the pre-cracking size was 2.0 mm and was machined by an electric discharge machining (EDM) technique to introduce a sharp crack tip starter for the BM, WM, and HAZ regions. In the welded joint, sharp pre-cracks for the HAZ were taken to conform to one of the HAZ locations. Load-line displacement was measured using a linear gauge assembly attached to the specimen, and the crack length was determined using a direct current potential drop (DCPD) technique, as shown in Fig. 5.1. Crack extension data were continuously collected using a data acquisition system. All of the experimental procedures followed the recommendations of the ASTM standard E1457. After the CCG testing, the CT specimens were broken open at liquid nitrogen temperature to measure the actual crack length. The actually measured final crack length (a_{mf}) was calculated from measurements made on the fracture surface at nine equally spaced points (so-called “nine points method”) using the enlarged photo of the fractured surfaces, because the individual measurements on the fracture surface vary due to crack front irregularities.

In addition, the predicted crack length (a_p) by DCPD technique was calculated by Johnson's formula:

$$\frac{a_p}{W} = \frac{2}{\pi} \cos^{-1} \left[\frac{\cosh(\pi Y_o / 2W)}{\cosh \left[\frac{V}{V_o} \cosh^{-1} \left\{ \frac{\cosh \pi Y_o / 2W}{\cosh \pi a_o / 2W} \right\} \right]} \right] \quad (5.1)$$

where, a_o = initial crack size (reference crack size for the reference voltage V_o), Y_o = the half distance between the output voltage loads, V = the output voltage, and W = the width of specimen. The a_p was compensated by the potential measurement error and became the corrected crack length a :

$$a = \frac{a_{mf} - a_o}{a_{pf} - a_o} (a_p - a_o) + a_o \quad (5.2)$$

where, a_{pf} is the predicted value of the final crack length. Hence, the difference values between the a_{mf} and a_{pf} were investigated whether their values were fairly included within the requirements of ASTM E1457, or not. Namely, the CCG data are valid for further processing if

$$0.85 \leq (\Delta a_{pf} / (a_{mf} - a_o)) \leq 1.15 \quad (5.3)$$

where, the predicted crack extension, Δa_{pf} , was calculated by substrating the initial crack length, a_o , from the finally predicted crack length, a_{pf} [11]. From these procedures, the validity for the CCG tests were identified for all the tested CT specimens at 550 °C and 600 °C.

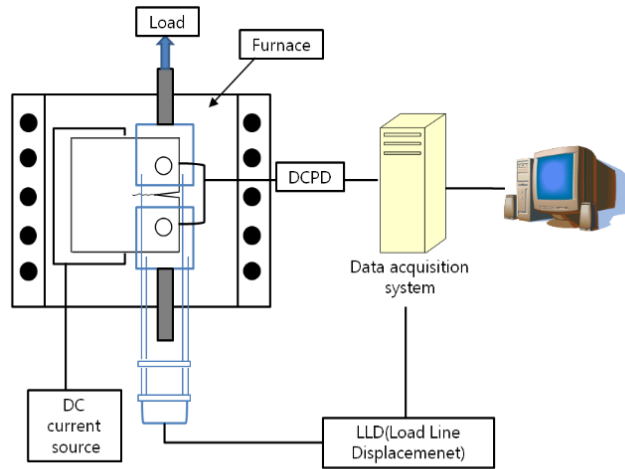


Fig. 5.1 A schematic diagram for the CCG tests.

5.1.2 Determination of material constants at 550°C and 600°C

The variations of micro-Vickers hardness for the BM, WM and HAZ regions in the welded joint were measured, as shown in **Fig. 5.2**. The thickness of the HAZ region was around 3.0~5.0mm. The hardness of the WM region shows higher values than that of the BM and HAZ regions, which means overmatch welding. It should be noted that the hardness in the HAZ decreases transitionally between the BM and WM, which might function as metallurgical notch. This HAZ should be very weak part with more possibility of cracking compared with the parts of BM and WM, and so-called Type-IV cracking may occur in this fine-grained zone which is located in the HAZ adjacent to the parent metal. It is known that the Type-IV cracks are parallel or offset from the fusion line.

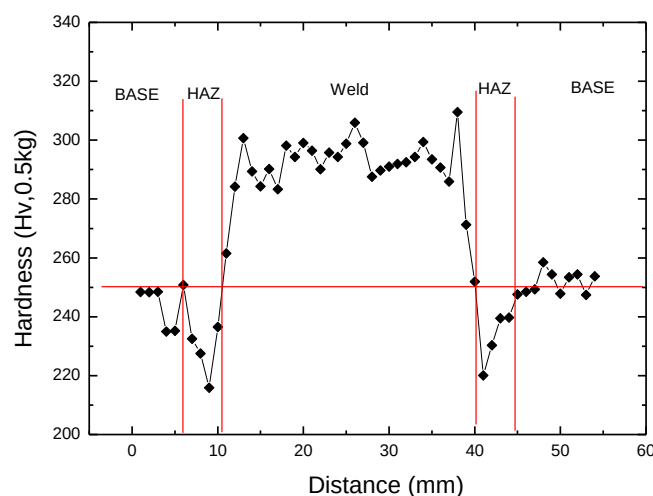


Fig. 5.2 Vickers micro-hardness for BM, WM, and HAZ of Gr.91 steel.

To calculate the fracture parameter of C^* , it is necessary to determine the D and m constants for the BM, WM, and HAZ. All values for the BM, WM, and HAZ were obtained from the tension tests at 550 °C and 600 °C.

Plastic constants were obtained in terms of $\varepsilon_p = D(\sigma/\sigma_y)^m$, where ε_p is true plastic strain, σ is true stress, and σ_y is yield stress. The relationship between plastic strain and true stress showed good linearity. From this linear relation, the D and m values for the BM, WM, and HAZ could be obtained at 550°C and 600°C. In addition, the creep constants of A and n were also obtained by a series of the creep tests for the BM, WM, and HAZ at 550°C and 600°C.

Fig. 5.3 and **Fig. 5.4** show the relationships between steady state creep rate and applied stress, which showed good linearity based on Norton's power law. At both temperatures of 550°C and 600°C, the WM and HAZ were higher in A and n values than the BM, and the HAZ is transitionally changed between the BM and WM. However, the HAZ is almost similar to the WM in the creep strain rate. The creep strain rates of the WM and HAZ was found to be the difference of approximately one order compared to the BM. It is believed that the WM and HAZ will be significantly attributed to a faster crack propagation rate. Accordingly, through the tension and creep tests at 550°C and 600°C, the material constants of D , m , A , and n were obtained for the BM, WM, and HAZ, respectively. A summary of these material constants at 550°C and 600°C, which were used to calculate the C^* fracture parameter, is given in **Table 5.1** and **Table 5.2**, respectively.

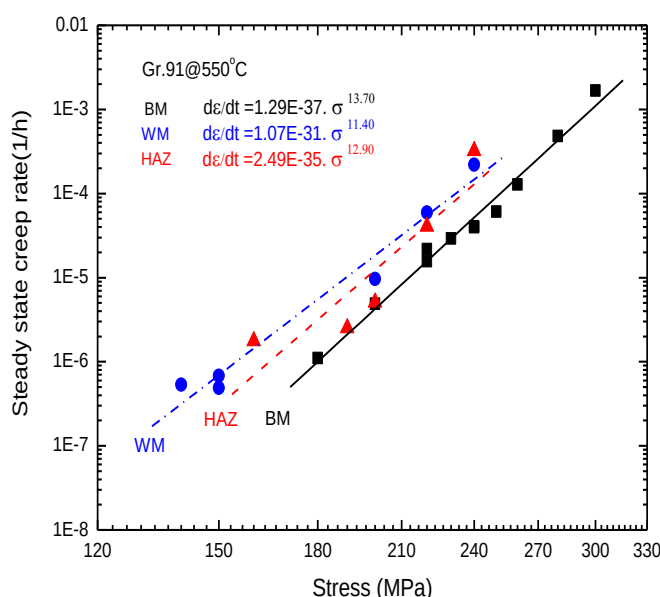


Fig. 5.3 Comparison of the steady state creep rate for the BM, WM, and HAZ at 550°C.

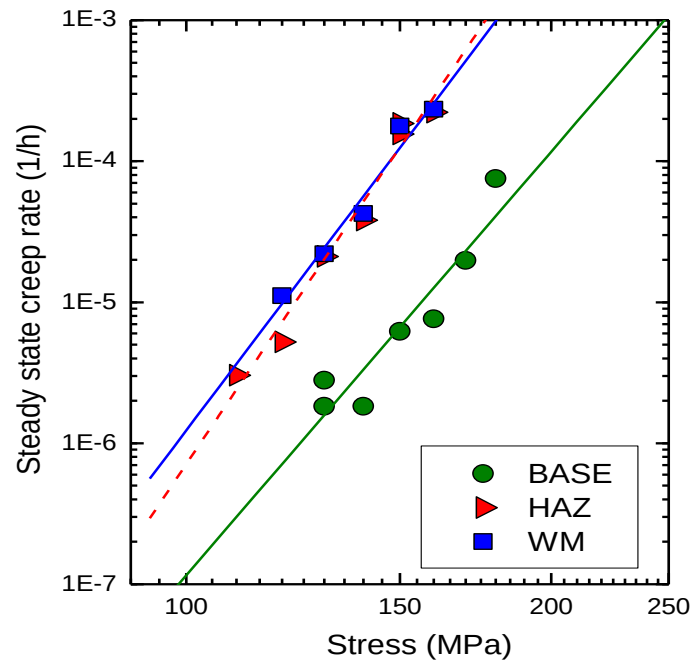


Fig. 5.4. Comparison of the steady state creep rate for the BM, WM, and HAZ at 600°C

Table 5.1. A summary of material constants obtained for the BM, WM, and HAZ at 550°C.

Material	σ_y (MPa)	$\varepsilon_p = D(\sigma / \sigma_y)^m$		$\dot{\varepsilon}_s = A\sigma^n$	
		D	m	$A(\text{MPa}^{-n}\text{h}^{-1})$	n
BM	338	1.23E-03	15.45	1.29E-37	13.70
WM	376	1.86E-03	13.55	1.07E-31	11.40
HAZ	321	1.76E-03	12.44	2.49E-35	12.90

Table 5.2. A summary of material constants obtained for the BM, WM, and HAZ at 600°C

Material	σ_y (MPa)	$\varepsilon_p = D(\sigma / \sigma_y)^m$		$\dot{\varepsilon}_s = A\sigma^n$	
		D	m	$A(\text{MPa}^{-n}\text{h}^{-1})$	n
BM	247.5	0.0017	6.4	1.28E-27	9.98
WM	298.2	0.00191	12.99	2.28E-29	11.27
HAZ	282.5	0.00184	11.13	3.26E-32	12.63

5.1.3 Calculation of C* fracture parameter

The fracture mechanics parameter, C*-integral, has been widely used to characterize the creep crack growth rate in metals undergoing a steady state creep. The steady state creep rate can be written using Norton's power law:

$$\dot{\epsilon}_s = A\sigma^n \quad (5.4)$$

The general form between the creep crack growth rate (da/dt) and the C^* can be expressed by

$$da/dt = B[C^*]^q \quad (5.5)$$

where n is the creep exponent, and the B and q coefficients are material constants, which are generally obtained from a regression line of the CCGR data. They are related to the intercept and slope, respectively, of the da/dt vs. C^* relationship on a log-log plot. To calculate the da/dt in Eq. (5.5), material data obtained in **Table 5.1** were used to calculate the C^* values. The relationship between da/dt and C^* was obtained for all samples.

In the CT specimen, the C^* value was calculated using Eq. (5.6), and load-line displacement rate ($\dot{v}_c$) due to creep strain was calculated by Eq. (5.8).

$$C^* = \frac{P\dot{v}_c}{B_N W} \eta\left(\frac{a}{W}, n\right) \quad (5.6)$$

$$\eta\left(\frac{a}{W}, n\right) = \frac{1}{(1-a/W)} \frac{n}{n+1} \left(\gamma - \frac{\delta}{n}\right) \quad (5.7)$$

$$\dot{v}_c = \dot{\epsilon}_c \frac{\phi B_N}{P} \left(\frac{2K^2}{E'} + (m+1)J_p \right) \quad (5.8)$$

where P is applied load, a is crack size, W is width of the specimen, $\dot{v}_c$ is total load-line displacement rate, B_N is net thickness of specimen, E' is elastic modulus for plane strain, K is stress intensity factor, $\dot{\epsilon}_c$ is crack growth rate, and m is stress exponent in the Ramberg-Osgood stress versus strain relationship.

5.1.4 Construction of CCG laws at 550 °C

Fig. 5.5 shows the plots of C^* vs. da/dt obtained for the BM (a), WM (b), and HAZ (c) of Gr.91 steel at 550°C. The solid lines show the regression curves obtained using the least squares fit method for each CCG data. From the regression lines, the creep crack growth laws for the BM, WM, and HAZ can be defined, as follows:

$$da/dt = 9.33 \times 10^{-3} (C^*)^{0.77} \quad (\text{BM}) \quad (\text{valid range : } 0.05 < C^* < 3.05 \text{ N/mm.h}) \quad (5.9)$$

$$da/dt = 2.63 \times 10^{-2} (C^*)^{0.82} \quad (\text{fWM}) \quad (\text{valid range : } 0.013 < C^* < 0.60 \text{ N/mm.h}) \quad (5.10)$$

$$da/dt = 1.93 \times 10^{-2} (C^*)^{0.91} \quad (\text{HAZ}) \quad (\text{valid range : } 0.02 < C^* < 3.27 \text{ N/mm.h}) \quad (5.11)$$

When the three CCGR curves for the BM, WM, and HAZ were overlapped together, it is shown that the WM and HAZ are faster than BM in terms of creep crack growth rate as shown in **Fig. 5.5(d)**. The WM is about 2.8 times faster than the BM. The CCGR line of the HAZ is transitionally changed and positioned between the WM and BM. The HAZ is thought to be complicated with the inhomogeneity among those materials and it affects the state of stress field or strain rate field near the weld fusion line and near the BM/HAZ interface, as a transition region between the BM and WM. It is assumed that the CCGR in the creep crack growth tests was similar to the creep strain rate in the creep tests.

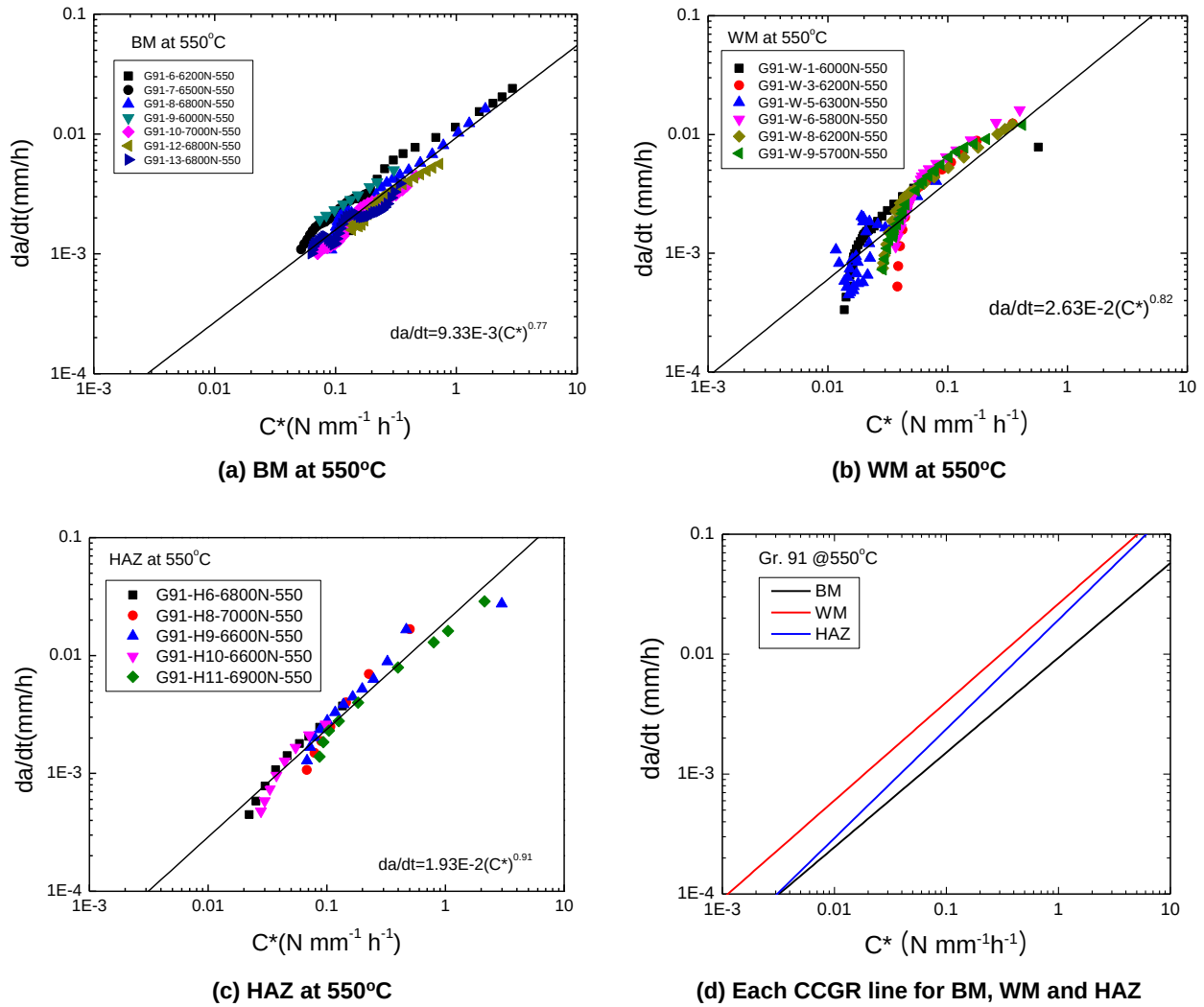


Fig. 5.5. Plot of C^* vs. CCGR lines for the BM, WM, and HAZ at 550°C

5.1.5 Construction of CCG laws at 600° C

The CCGR laws at 600°C were constructed similar to the case of 550°C. **Fig. 5.6** shows the plots of C^* vs. da/dt obtained for the BM(a), WM(b), and HAZ(c) of Gr.91 steel at 600°C. The solid lines show the regression curves obtained using the least square fit method for each CCG data. The creep crack growth laws for the BM, WM and HAZ were defined, as follows:

$$da/dt = 1.89 \times 10^{-2} (C^*)^{0.77} \quad (\text{BM}) \quad (\text{validity range : } 0.003 < C^* < 1.5\ \text{N/mm.h}) \quad (5.12)$$

$$da/dt = 3.62 \times 10^{-2} (C^*)^{0.85} \quad (\text{WM}) \quad (\text{validity range : } 0.025 < C^* < 2.5\ \text{N/mm.h}) \quad (5.13)$$

$$da/dt = 3.71 \times 10^{-2} (C^*)^{0.86} \quad (\text{HAZ}) \quad (\text{validity range : } 0.012 < C^* < 4.0\ \text{N/mm.h}) \quad (5.14)$$

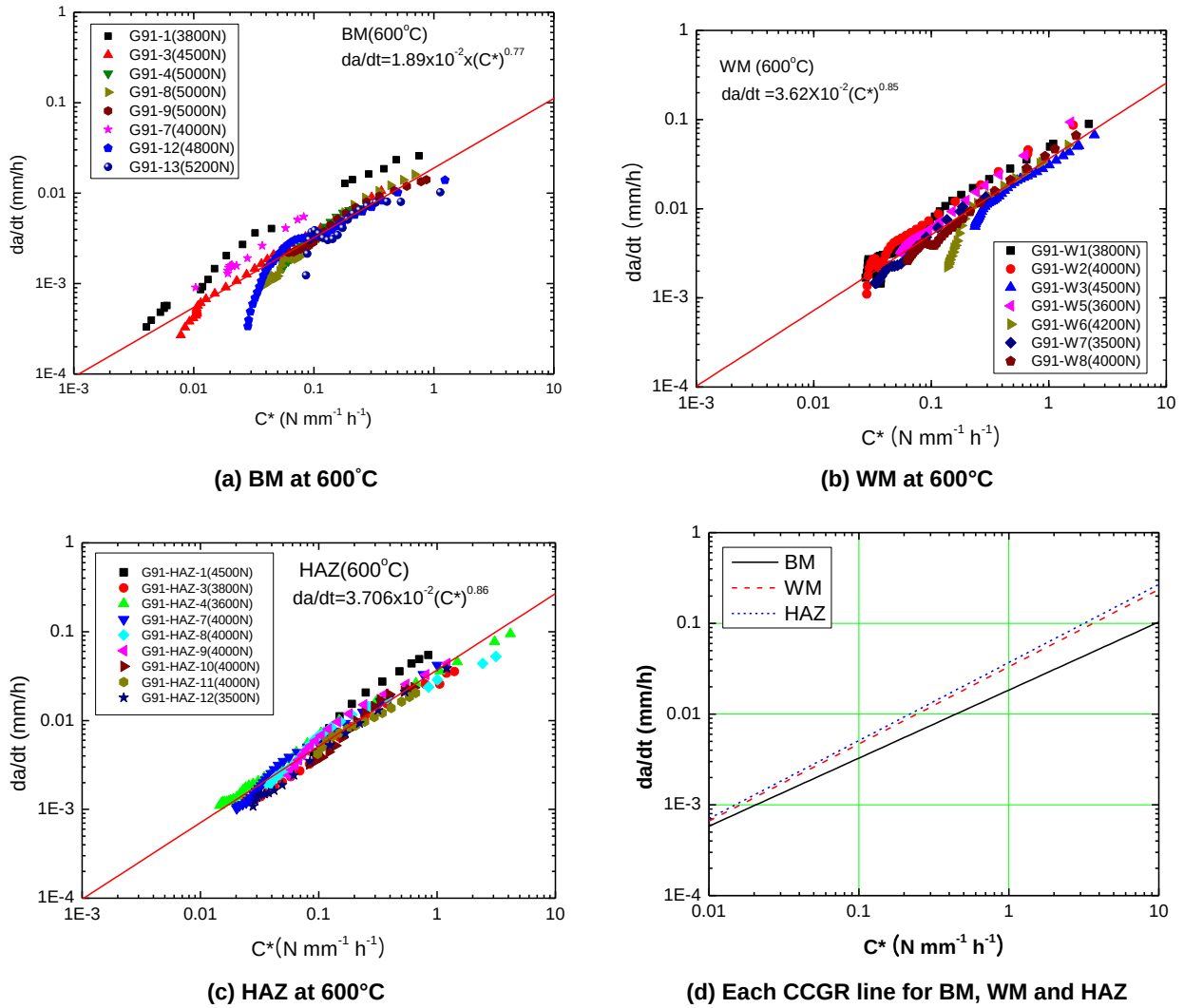


Fig. 5.6. Plot of C^* vs. CCGR for the BM, WM, and HAZ

5.2 High temperature material strength tests for Gr.91 steel

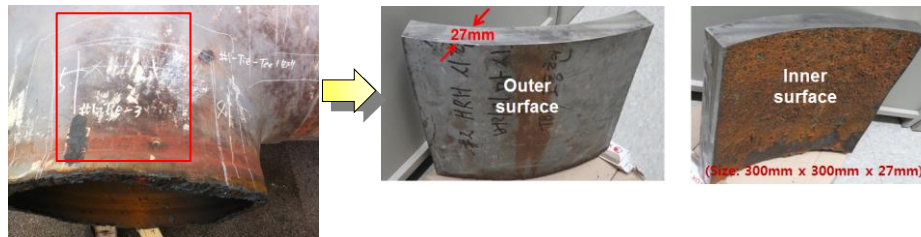
5.2.1 Virgin Gr.91 Specimens

The chemical compositions of Gr.91 steels in the ETD codes for Gr.91 steel plate, forged Gr.91(F91) steel in the ASME code, unexposed original Gr.91 steel, and the KAERI virgin plate used for fabrication of the material specimens in the present study are given in **Table 5.3**. As shown in **Table 5.3**, the RCC-MRx code restricts the carbon range more strictly, and a higher carbon from 0.06 to 0.08 would improve the creep strength, and decreasing the upper limit from 0.15 to 0.12 would increase the microstructure stability by the suppression of carbide coarsening and undesirable phase formation. The service-exposed Gr.91 steel was sampled from a tee fitting of forged Gr.91 (F91) steel in a USC plant. The chemical compositions were within the range of ASME F91 as shown in **Table 5.3**.

In the present study, virgin Gr.91 steel material specimens were made with the 'KAERI virgin plate' rather than the 'unexposed original' Gr.91 steel in **Table 5.3**, which is not from the same heat of the service-exposed Gr.91 steel. It should be ideal to use the same heat material with the unexposed original Gr.91 steel in material testing. In reality, however, it is hard to secure the original material and even the mill sheet because it is either confidential or not possible to obtain. It is expected that the differences in the chemical compositions of the original batch and KAERI's batch should not result in big differences in the tensile strengths and J-R curves.

Table 5.3. Chemical compositions of Grade 91 steel (wt.%)

Code/Test	C	Mn	P	S	Si	Ni	Cr	Mo	V	Al	Others
ASME (Gr.91 plate)	0.06-0.15	0.25 - 0.66	0.025	0.012	0.18 - 0.56	0.43	7.90 - 9.60	0.80 - 1.10	0.025 -0.08	0.02	
RCC-MRx (Gr.91 plate)	0.08-0.120	0.30 - 0.60	≤0.02	≤0.005	0.20 - 0.50	≤0.20	8.00 - 9.50	0.85 - 1.05	0.03 - 0.07	≤0.04	Nb 0.06 - 0.10
ASME (F91)	0.08-0.12	0.3-0.6	0.025 max	0.025 max	0.2-0.5	0.4 max	8.0-9.5	0.85-1.05	0.18-0.25	Al 0.02	Cb 0.03-0.07 Ti 0.01 Zr 0.01
Unexposed, original (Gr.91)	0.08	0.38	0.018	0.003	0.34	0.29	8.75	0.875	0.242	0.014	Cb 0.079
KAERI Virgin plate (Gr.91)	0.115	0.415	0.012	0.0014	0.23	0.22	8.9	0.869	0.051	0.014	Cb 0.079 N 0.038

**Fig. 5.7 Gr.91 tee junction and sampled part after 73,716 h of service at a USC plant**

5.2.2 Service-exposed Gr.91 steel

It is known that Gr.91 steel degrades more severely under long-time service at high temperature compared with austenitic stainless steel 316. Therefore, the effects of thermal aging were investigated, and a comparison was made regarding the strength reduction factors from long-time service in the ASME-NH. Thermally aged Gr.91 steel was used to investigate the long-time service effects and to quantify the conservatism of the material strength properties in the ETD (elevated temperature design) codes. The service-exposed Gr.91 steel was sampled from a tee junction in the reheat steam piping system of a USC thermal power plant in Korea with an accumulated operation time of 73,716 hours (~8.4 years). The images of the tee fitting and sampled plate are shown in **Fig. 5.7**. The forged tee fitting had an inner diameter of 760 mm and a thickness of 27 mm. The design temperature and pressure of the piping system were 574°C and 0.54 MPa, respectively, while the operating temperature and operating pressure were 569°C and 4.67 MPa, respectively.

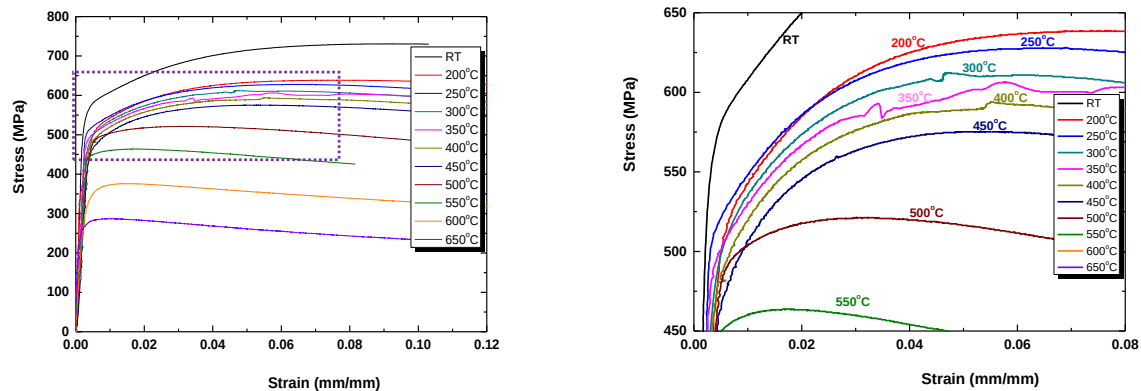
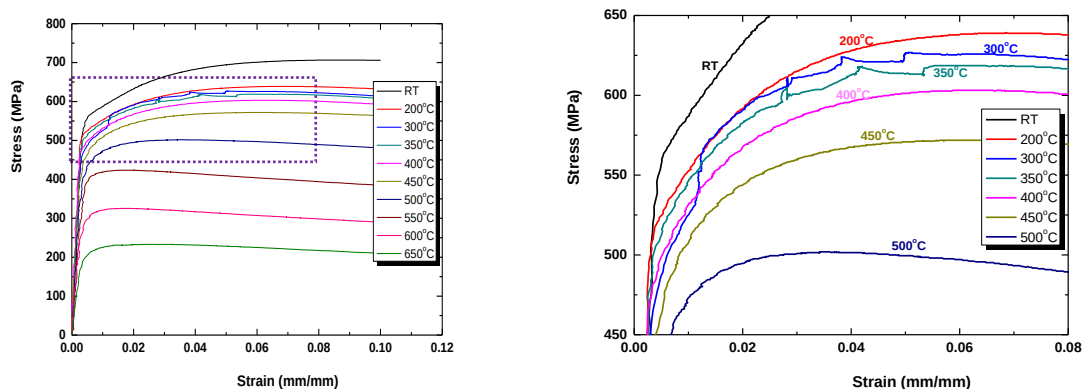
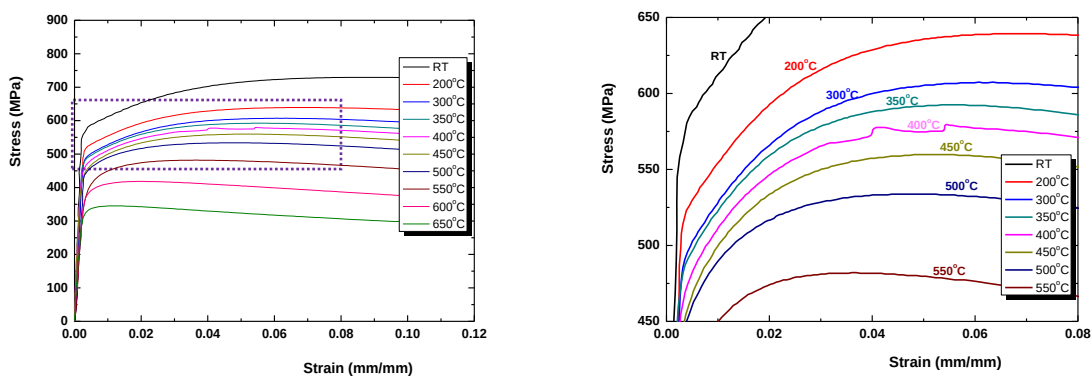
The maximum stress intensity at the tee fitting under pressure and thermal loading under normal operating conditions was evaluated as 79.1MPa according to finite element analysis, which is not a high stress level under combined thermal and mechanical loads.

5.2.3 Material strength tests

A number of tension tests were conducted for the virgin Gr.91 steel specimens and serviced-exposed Gr.91 steel specimens based on the test matrix in **Table 5.4**. As shown in the test matrix, tensile tests at 10 temperatures from room-temperature to 650 °C were conducted both for the virgin and service-exposed Gr.91 steel specimens. The strain rate of the test for service-exposed Gr.91 steel was set as 1 mm/min at the cross head, which is equivalent to a strain rate of 6.67×10^{-4} (1/s) at the gauge part of the specimen, while the strain rates of virgin Gr.91 steel were 6.67×10^{-3} , 6.67×10^{-4} , 6.67×10^{-5} as shown in **Table 5.4**. Since the size of the sampled plate for service-exposed Gr.91 steel was relatively small, the number of the specimens for strength tests and J-R tests was limited, which is why tension tests were conducted for one strain rate only.

Table 5.4. Test matrix of the tension tests

Material	Strain rate (1/s)	Test temperatures (°C)
Virgin Gr.91	6.67×10^{-3}	RT, 200, 300, 350, 400, 450, 500, 550, 600, 650
	6.67×10^{-4}	RT, 200, 300, 350, 400, 450, 500, 550, 600, 650
	6.67×10^{-5}	RT, 200, 300, 350, 400, 450, 500, 550, 600, 650
Service-exposed Gr.91	6.67×10^{-4}	RT, 200, 300, 350, 400, 450, 500, 550, 600, 650

Fig. 5.8 Tension test results for strain rate of 6.67×10^{-5} (mm/mm/sec) virgin & agedFig. 5.9 Tension test results for strain rate of 6.67×10^{-4} (mm/mm/sec) virgin & agedFig. 5.10 Tension test results for strain rate of 6.67×10^{-3} (mm/mm/sec) virgin & aged

A number of tension tests for Gr.91 steel were conducted over a range of strain rates (6.67×10^{-3} , 6.67×10^{-4} , and 6.67×10^{-5}) temperatures (from room temperature and 650°C) for virgin Gr.91 steel and service-exposed Gr.91 steel with the strain rate of 6.67×10^{-4} as shown in **Table 5.4**.

The test results for the strain rate of 6.67×10^{-3} , 6.67×10^{-4} , and 6.67×10^{-5} are shown in **Fig. 5.8**, **5.9** and **5.10**, respectively. As shown in **Fig. 5.8**, ratchet type serration occurred at the temperature of 300°C and 350°C for the strain rate of 6.67×10^{-5} . In addition, serration occurred at the temperature of 300°C, 350°C and 400°C for the strain rate of 6.67×10^{-4} , and 400°C for the strain rate of 6.67×10^{-4} . The stress-strain test results show that the serration temperature tends to increase as the strain rate increases for Gr.91 steel.

For the service-exposed Gr.91 steel, the tension test results also showed serrated flow as shown in **Fig. 5.11**. Serration flow in service-exposed Gr.91 steel was more severe than that of virgin Gr.91 as shown in **Fig. 5.9** and **Fig. 5.11**.

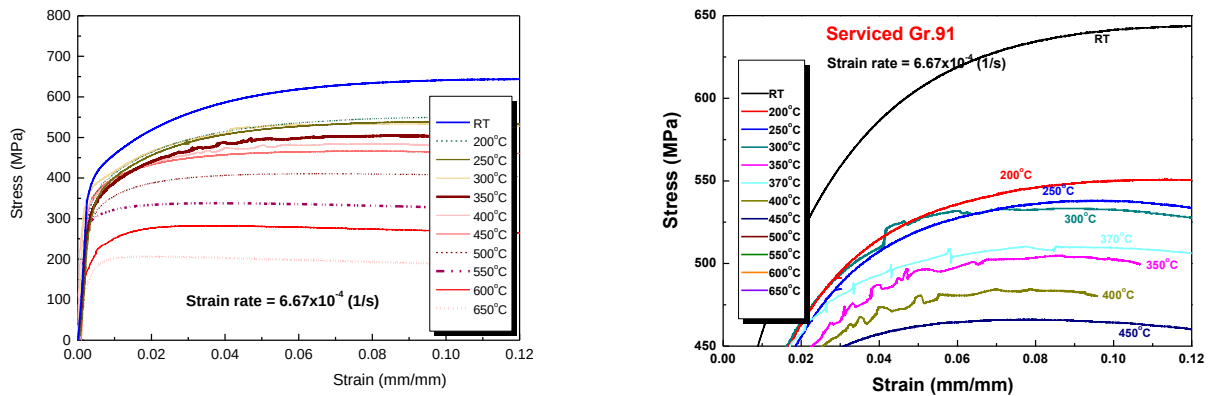


Fig. 5.11 Tension test results for service-exposed Gr.91 steel (73,716h @569°C, 46.7bar) with 6.67×10^{-4} (mm/mm/sec) virgin & aged

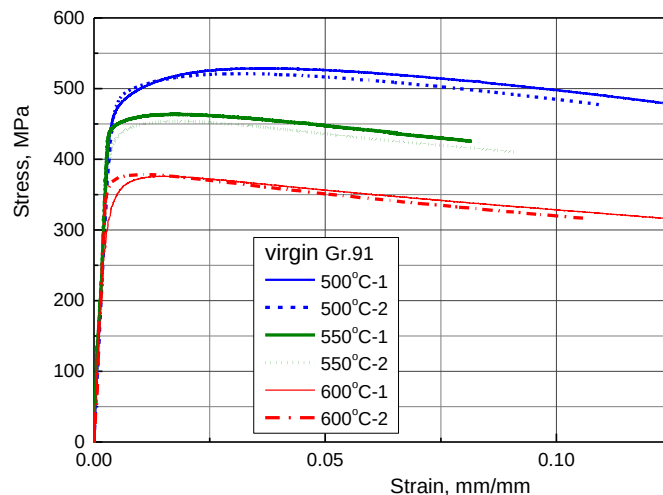


Fig. 5.12 Tension test results with two test specimens at 500°C, 550°C and 600°C

All tension tests at a specific case in Table 5.4 were conducted with only one specimen but duplicate tests were conducted over the temperatures of 500 °C, 550 °C and 600 °C to investigate the data scatter. The duplicate test results showed very similar behavior with the maximum data scatter 1.51% for 500 °C, 3.88% for 500 °C and 0.61% for 600 °C as shown in Fig.5.12. The stress and strain in Fig. 5.12 are engineering

values. The discrepancies in the duplicate tests were reasonably small, so other tests were conducted with one specimen each.

5.2.4 Comparison of tension test results for virgin and service-exposed Gr.91 steel

A comparison of the two test results for the same strain rate (6.67×10^{-4} mm/mm/sec) shows that the tension curve steel dropped for the aged compared to virgin by 29.2% at room temperature, 22.8% at 300 °C, 33.0% at 500 °C, and 36.2% at 600 °C as shown in **Fig. 5.13** to **Fig. 5.15**.

The elongation behavior of the Gr.91 steel before and after long-time service was investigated over a range of the temperatures. The results show that elongation increased as much as 17.8% to 23%, as shown in **Fig. 5.16**, due to long-time service at high temperature.

Ferritic-martensitic steels, such as Gr.91 steel, owe their creep resistance to the effects of a fine precipitation-stabilized dislocation network formed during the tempering of martensite. The strength reduction of the service-exposed Gr.91 steel is linked with increase in the lath width and the reduction in the dislocation density due to the recovery and coarsening of the grain boundary carbides under long-time service.

However, ductility increased due to thermal aging at high temperature. Elongation properties are provided in the RCC-MRx only for austenitic stainless steel of 316L(N) and 316L but not for Gr.91 steel.

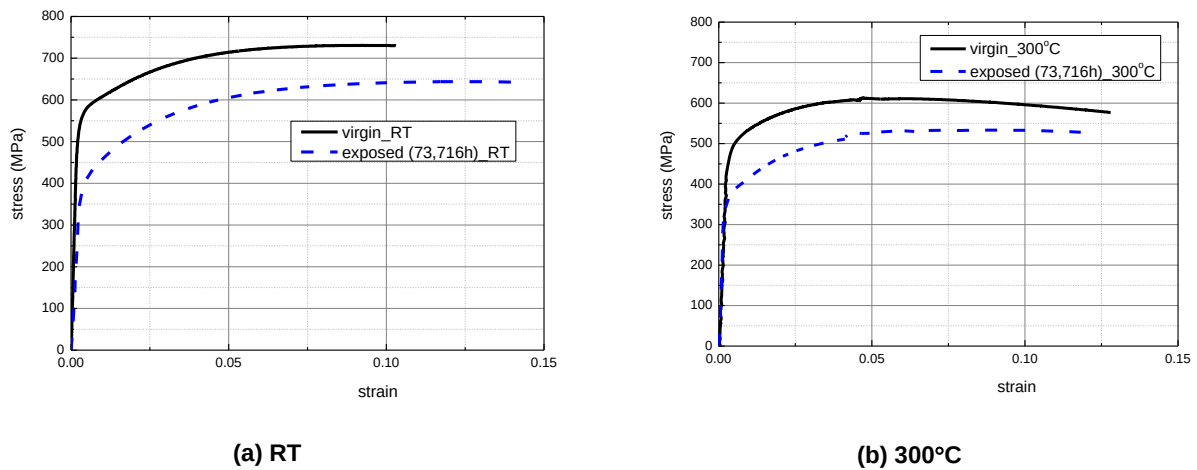


Fig. 5.13 Comparison of the tensile test results of the virgin and service exposed Gr.91 steels at RT and 300°C

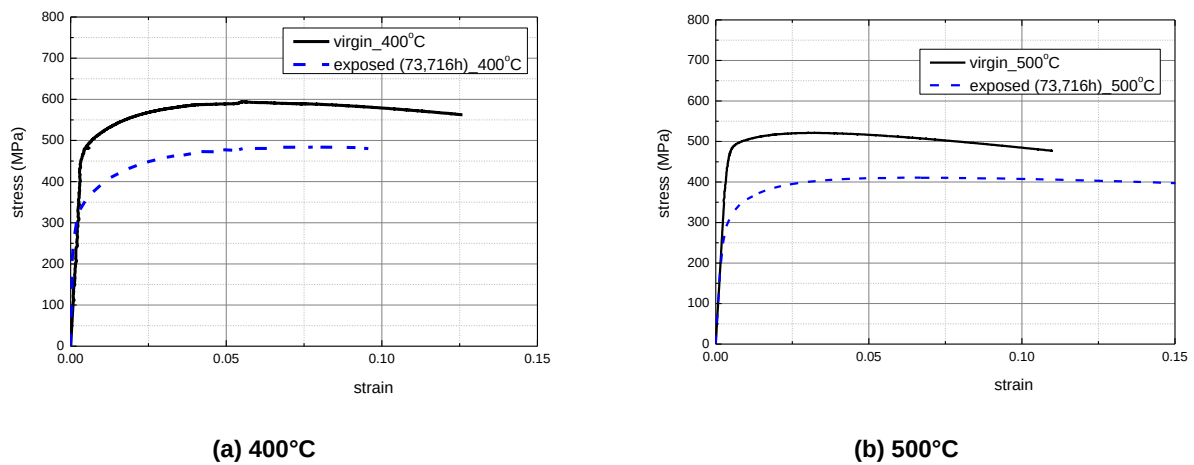
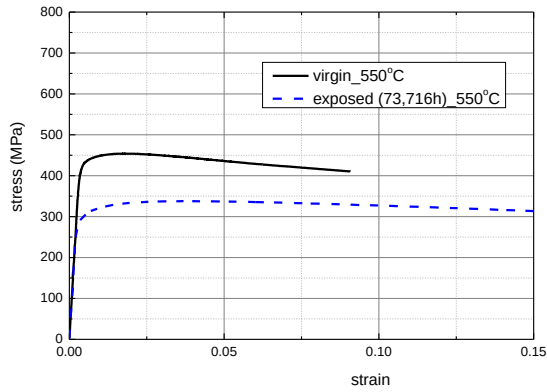
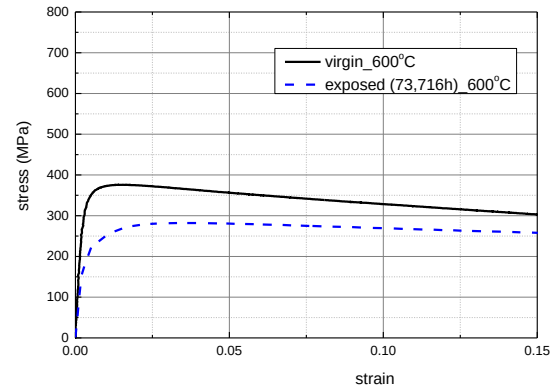


Fig. 5.14 Comparison of the tensile test results of the virgin and service exposed Gr.91 steels at 400°C and 500°C



(a) 550°C



(b) 600°C

Fig. 5.15 Comparison of the tensile test results of the virgin and service exposed Gr.91 steels at 550°C and 600°C

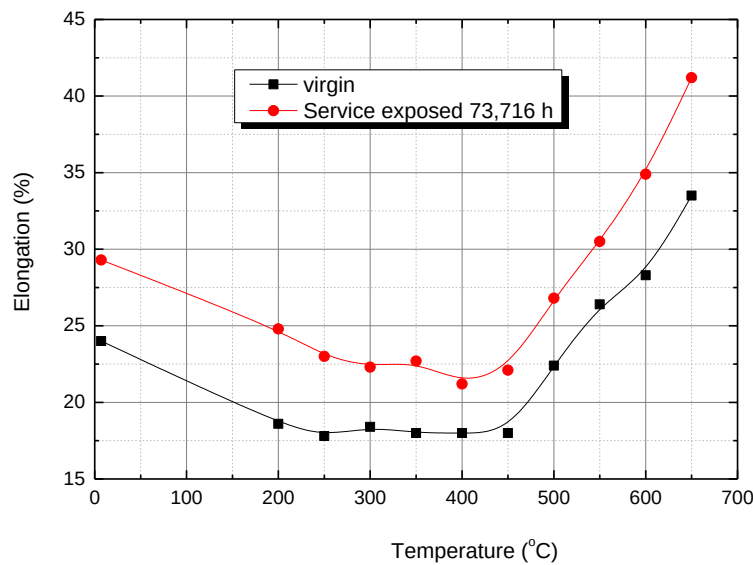


Fig. 5.16 Comparison of elongation between the virgin Gr.91 and service-exposed Gr.91

5.3 J-R tests for Gr.91 steel

Fracture toughness is an important factor along with material strength in maintaining the integrity of high-temperature components. The fracture properties for austenitic stainless steel are provided relatively well in the RCC-MRx compared with those of Gr.91 steel. It was reported that there are issues on non-conservatism on the property of creep crack growth (CCG) for Gr.91 steel in the RCC-MRx Tome 6 based on CCG tests.

The other important issue for Gr.91 steel is that the J-R properties are not provided in A9 of the RCC-MRx which are required for leak before break (LBB) analysis. This study was intended to provide the J-R properties of Gr.91 steel for a range of temperatures. In addition, the effect of long-time service on the J-R behavior in Gr.91 steel was investigated.

5.3.1 J-R test details

Table 5.5. Test matrix of the J-R fracture tests

Material	Test temperature (°C)	Cross head speed (mm/min)
Virgin Gr.91 steel	300, 370, 400, 425, 500, 550, 600	1.0
Service exposed Gr.91 steel	300, 370, 400, 425, 500, 550, 600	

The J-R fracture specimens for the virgin Gr.91 and service-exposed Gr.91 steels were standard ASTM compact tension specimens with a thickness of 25.4 mm and a chevron notch.

The J-R fracture tests were conducted according to the procedure of ASTM E1820 with a fatigue pre-crack larger than 1.3 mm at the notch tip and longer than 0.05 times the notch length.

As part of investigating the fracture toughness of Gr.91 steel, the J-R fracture curves for the virgin and service-exposed Gr.91 steel specimens were obtained for a range of temperatures, and their J-R behavior was investigated independently and compared to quantify the effect of long-time service at high temperature.

As shown in the J-R test matrix of **Table 5.5**, the tests were performed at 8 temperatures from room temperature to 600°C, and the strain rate of the J-R test at cross-head was 1mm/min.

5.3.2 Comparison of J-R test results for virgin and service-exposed Gr.91 steel specimens

The J-R fracture curves obtained for the eight temperatures are shown in **Fig. 5.17**. The J-R curve has the minimum J value at 400°C up to a crack extension of 5 mm, which is a unique behavior in that each J resistance at 500°C, 550°C and 600°C is greater than that at 400°C for the virgin Gr.91 steel shown in **Fig. 5.17**.

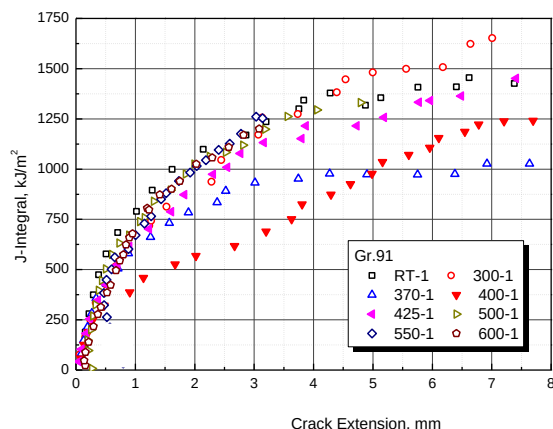


Fig. 5.17 J-R curve of virgin Gr.91 steel

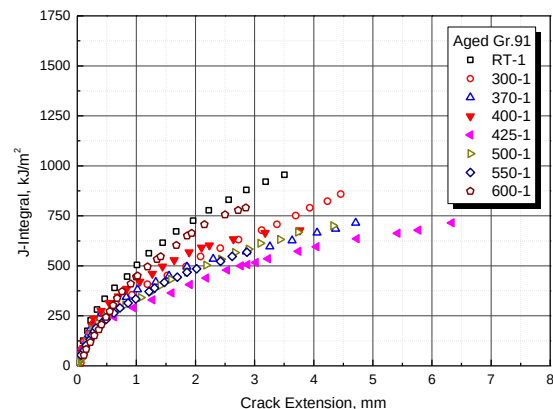


Fig. 5.18 J-R curve of service-exposed (73,716 h) Gr.91 steel

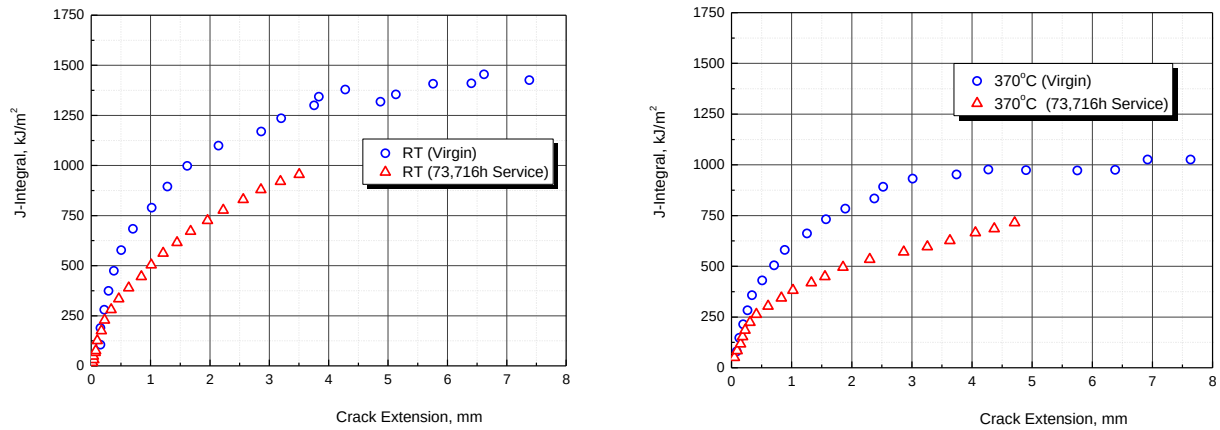


Fig. 5.19 Comparison of the J-R curves for the virgin and service-exposed Gr.91 steels at RT and 370°C

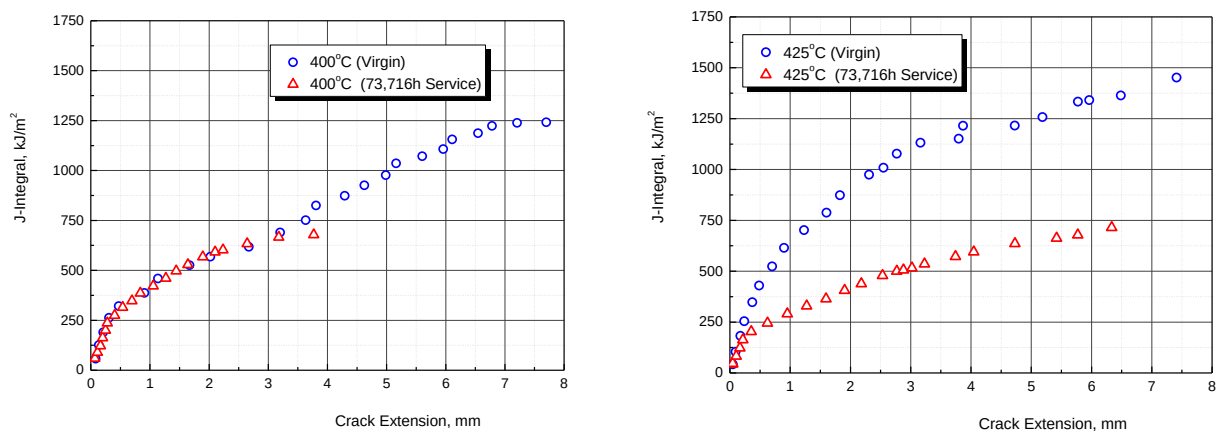


Fig.5.20 Comparison of the J-R curves for the virgin and service-exposed Gr.91 steels at 400°C and 425°C

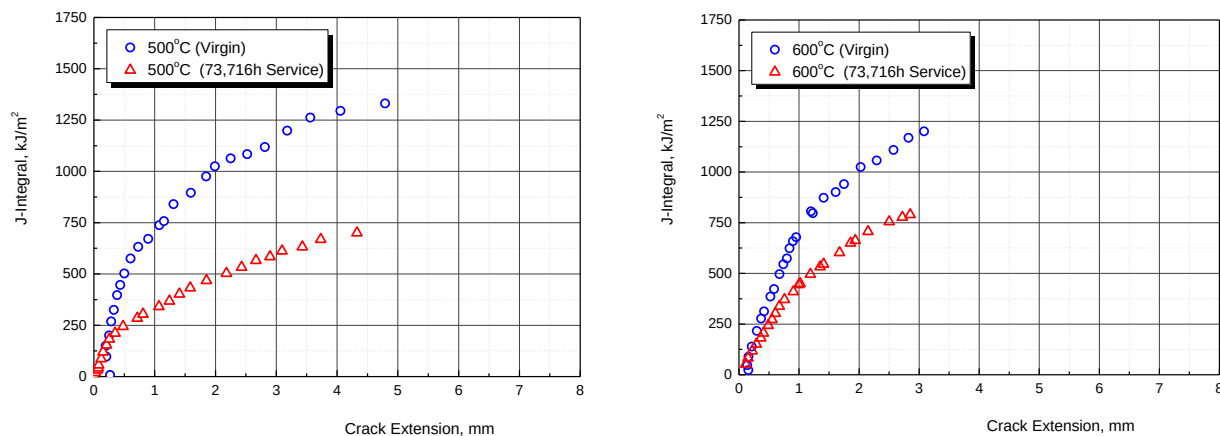


Fig. 5.21 Comparison of the J-R curves for the virgin and service-exposed Gr.91 steels at 500°C and 600°C

In the case of the service-exposed Gr.91 steel, the J-R curve for 425 °C was the lowest as shown in **Fig. 5.18**, which means that the fracture toughness minimum temperature changed from 400°C (virgin Gr.91) to 425°C (service-exposed Gr.91). In addition, the magnitude of the J-integral decreased due to long-time service. The J value dropped as much as 51% for 425°C and 16% for 400°C due to long-time service, which are significant reductions in fracture resistance.

The J-R curves before and after service are compared in **Figs. 5.19 to 5.21** to see the effects of long-time service on Gr.91 steel. As shown in the figures, the J-R curve dropped as much as 25.5% at room temperature, 38.1% at 370 °C, 17.7% at 400 °C, 52.9% at 425 °C, and 38.5% at 600 °C.

6 Concluding Remarks

This report contains three separate studies focussing on ferritic-martensitic Gr 91 steels but also covering 316L austenitic steels:

- The largest effort was on creep-fatigue crack growth tests and assessment following the ASTM Standard E2760-16 with the objective to quantify the effect of hold time for different loads and temperature but also to assess the relatively recent test standard (JRC, CVR and CIEMAT)
- An experimental simulation and numerical assessment of thermal fatigue for pipe components in order to have a load case representative of components (JRC);
- An assessment of creep crack growth, tensile and fracture properties for Grade 91 welds and service exposed material (KAERI).

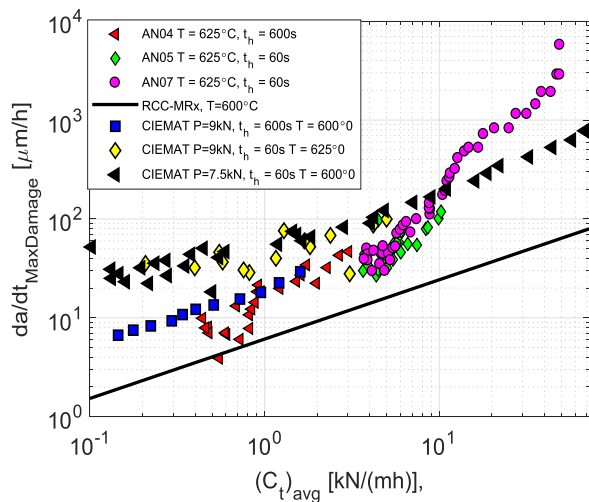


Figure 6.1a Comparison of versus crack growth rate for the 60 mm JRC plate with hold time of 60 and 600 seconds. (JRC and CIEMAT) (Figure 3.2.13a and 3.3.6)

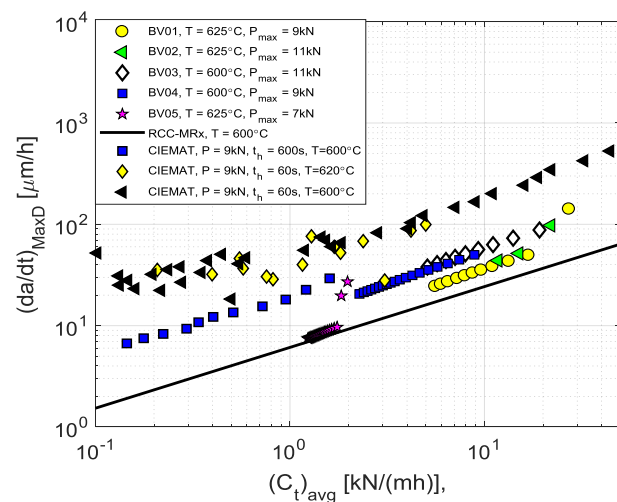


Figure 6.1b Comparison of versus crack growth rate for the 60 mm JRC plate. JRC block load 60 s hold time CIEMAT 60 and 600 seconds hold time (Figure 3.2.17c and 3.3.6)

The creep-fatigue tests are more complex than fatigue crack growth tests or creep crack growth tests. The importance of very accurate measurement of crack length and load-line displacement during the tests to derive the creep properties was demonstrated. It is generally not possible to measure these quantities for each cycle so some kind of smoothing is required. The 60 mm Grade 91 plate provided by JRC was used by all three partners. CIEMAT had generally less noise in their results than JRC, partly due to longer crack growths for the determination of crack growth rates but also more stable and reliable recording of the load-line displacement. **Figure 6.1a** shows the crack growth rate versus $(C_t)_{avg}$ for JRC and CIEMAT from the experiments. Creep-fatigue data have a natural scatter so there globally a reasonable agreement except but the JRC data has a larger scatter. **Figure 6.1b** shows the JRC data for the block loading 60 seconds for dominant damage and $(C_t)_{avg}$ by the analytical formula 3.5 together with the CIEMAT data. The analytical JRC actually has a better agreement than in **Figure 6.1a** and without the up-shift. Creep-fatigue data typically has a certain scatter between tests. CIEMAT participated in the round robin-test in support of the development of the ASTM standard whereas CVR and JRC had had not performed these tests prior to MATISSE. Additional tests are needed to derive scatter bands and further improve the test procedures.

The thermal fatigue tests by JRC represented more complex loadings cracking representative for in-service conditions. For the assessment of creep-fatigue crack growth the C-integral and crack growth rate cannot be measured directly and instead numerical models must be adopted, which requires an appropriate constitutive model. In particular for ferritic-martensitic steels this becomes very complicate due to cyclic softening and associated increase of creep rates. However in this case the creep contribution was negligible compared to fatigue due to the short duration of high tensile stresses induced by the thermo-mechanical loads. This indicates that crack propagation, at least at 550°C, is more a high-temperature fatigue problem.

In nuclear power plants fracture usually occurs in welds and the material properties will change during the service life, which is very important to take into account for the long-term integrity of components. It therefore important to have data that representative for welds, in particular the weakest part, and the possible degradation of material properties under in-service conditions. KAERI clearly demonstrated the reduction in fracture toughness of service exposed Gr 91 steel.

7 References

1. E2760-16 Standard Test Method for Creep-Fatigue Crack Growth Testing, ASTM International
2. Riedel, H., *Fracture at High Temperatures*. Materials Research and Engineering 1987, Berlin: Springer Verlag.
3. N.B. Adefris and A. Saxena, Creep-Fatigue Crack Growth Behavior in an Ex-service Cr-Mo-V Rotor Steel, Subcontract No. 06-1536, EPRI Research Project 2481-5
4. N. B. Adefris, D. L. McDowell and A. Saxena, An Alternative Analytical Approximation Of The Ct Parameter, *Fatigue & Fracture of Engineering Materials & Structures* 1998; 21: 375–385
5. P.S. Grover and A. Saxena, Modelling the effect of creep-fatigue interaction on crack growth, *Fatigue Fract Engng Mater Struct* 22, 1999, 111-122
6. S.B. Narasimhachary, A. Saxena, Crack growth behavior of 9Cr-1Mo (P91) steel under creep–fatigue conditions, *International Journal of Fatigue* 56 (2013) 106–113
7. Ashok Saxena, Creep and creep–fatigue crack growth, *Int J Fract* (2015) 191:31–51
8. B. Fournier et al Analysis of the hysteresis loops of a martensitic steel Part1: Study of the influence of strain amplitude and temperature under pure fatigue loadings using a an enhanced stress partitioning method, *Mat. Science Engineering*, A437, 2006, 183-196
9. B. Fournier et al Comparison of various 9-12%Cr steels under fatigue and creep-fatigue loadings at high temperature, *Mat. Science Engineering*, A528, 2011, 6394-6495
10. R. A. Ainsworth, R5 procedures for assessing structural integrity of components under creep and creep–fatigue conditions, *International Materials Reviews* 2006 VOL 51 NO 2, pp 107 – 126
11. H. Sehitoglu and W. Sun, The Significance Of Crack Closure Under High Temperature Fatigue Crack Growth With Hold Periods, *Engineering Fracture Mechanics* Vol. 33, No. 3, pp. 371-388, 1989
12. AFCEN, RCC-MRx Design and Construction Rules for Mechanical Components of Nuclear Installations 2012 Edition, 2012, AFCEN: Paris.
13. Karl-Fredrik Nilsson, Francesco Dolci, Thomas Seldis, Stefan Ripplinger, Aleksander Grah, Igor Simonovski, Assessment of thermal fatigue life for 316L and P91 pipe components at elevated temperatures, *Engineering Fracture Mechanics* 168 (2016) 73–91
14. Charlesworth J.P., Terple J.A.G., *Engineering Applications of Ultrasonic Time-of-Flight Diffraction Research Studies* 1989: John Wiley & Sons Inc, 1989.
15. Nilsson, K.-F., et al., Analysis of Crack Morphologies and Patterns from Thermal Fatigue Using X-ray Tomography. *Procedia Materials Science*, 2014. 3: p. 2180-2186.
16. Chaboche, J.L., *Time-independent constitutive theories for cyclic plasticity*. International Journal of Plasticity, 1986. 2(2): p. 149-188.
17. Chaboche, J.L., Constitutive equations for cyclic plasticity and cyclic viscoplasticity. International Journal of Plasticity, 1989. 5(3): p. 247-302