**T73S03 (R5V4/5/7) – Session 41:**
**Experimental Determination of Creep Crack Growth Laws**

Last Update: 23/5/16

*Obtaining CCG data; Basic experimental arrangement & instrumentation; LLD rate, DCPD; Growth calibration; Code size requirements; Empirical formulae for C*; Can C(t) be measured?; Typical da/dt v C* plots; Tails: why they occur & comparison with trend line; Is* $da/dt = A.C(t)^q$ *justified? Real data: puzzle over 316H HAZ; Typical range of experimental C* cf assessments of plant (extrapolation)*

Qu.: What form do creep crack growth laws take?

The most common growth law used in practice is,

$$\frac{da}{dt} = AC(t)^q \tag{1}$$

In some cases $C(t)$ may reduce to $C^*$.

Qu.: Is this always valid?

No.

In particular, R5 recommends this growth law only if the crack tip is outside the cyclic plastic zone (the region of hysteresis). If so then the corresponding fatigue crack growth law is a Paris Law, $\frac{da}{dN} = C\Delta K_{eff}^m$. The total crack growth is then simply the sum of the creep and fatigue crack growths.

Qu.: What about inside the cyclic plastic zone?

Inside the cyclic plastic zone the appropriate fatigue crack growth law is of the form $\left(\frac{da}{dN}\right)_f = B'a^Q$ and the total crack growth is given by,

$$\frac{da}{dN} = \left(\frac{da}{dN}\right)_f \left(1 - D_c^{surf}\right)^{-2} \tag{2}$$

where $D_c^{surf}$ is the surface creep damage. Hence, in this situation the effect of creep is to enhance the fatigue crack growth rather than causing growth directly. The reason is that the cyclic plasticity wipes out the crack tip field and hence $C(t)$ is irrelevant.

Qu.: What is the subject of this session?

Here we are interested only in the experimental derivation of growth laws of the form $\frac{da}{dt} = AC(t)^q$ - and in practice experiments are invariably interpreted as being in the $C^*$ regime, so the experimentally derived laws are really $\frac{da}{dt} = AC^{*q}$.

Qu.: How are the parameters $A$ and $q$ found experimentally?

The best source for an overview of creep crack growth testing procedures and their interpretation is Ref.[1] (McLennon & Allport), which leans upon Refs.[2, 3]. R5V4/5 Appendix A1 is also worth a look. However, unlike fracture toughness J-testing, there is no British Standard for ccg testing, so Refs.[1-4] are the most authoritative sources.

## Qu.: What specimens are used in ccg testing?

As for J-testing, the best specimen to use is the Compact Tension Specimen (CTS), preferably a "2 inch" specimen (i.e., of approximate dimensions 2 x 2 x 1 inches, though actually slightly bigger).

The CTS is neither a tension specimen nor particularly compact.

Ref.[1] considers 6 specimen types in addition to the CTS. These are,

DENT, CCP, SENT, 3PB, CCB, CST

Ref.[1] summaries the K solutions, eta-factor solutions, reference stress solutions, etc., for all these specimens. These are needed to interpret the data from such specimen tests.

The specimens CTS, CST, SENT and 3PB are all bending dominated (despite the word "tension" in the titles of the first three).

The specimens DENT, CCP and CCB are all membrane tension dominated.

## Qu.: Why are CTS tests preferable?

Two reasons are,

- The CTS is the most highly constrained of all the above specimens, and hence provides the most conservative result;
- Most past tests have used CTS and so a direct comparison is available uncomplicated by differences due to specimen geometry.

## Qu.: What are the drawbacks of using CTS?

Two drawbacks are,

- They are bulky specimens and full size ("2 inch") CTS are only rarely possible from ex-service material;
- They are likely to be more constrained than the plant application, and hence may be too onerous.

## Qu.: How are the cracks introduced?

After machining the basic specimen, the usual procedure is to introduce the crack by,

- Electro-discharge machining (EDM) a narrow slot (~0.2mm wide should be achievable);
- Fatiguing the specimen at a $\Delta K$ less than the intended creep test load so as to grow a sharp fatigue crack from the EDM notch.

The fatigue sharpening operation can be delicate and requires some experience to get right. Too much fatigue will result in too long a crack, which effectively ruins the specimen. Too little and there may be no fatigue sharpened crack, which might invalidate the test (due to the finite radius of the EDM slot). The snag is that you cannot be certain about the fatigue grown crack until you break the specimen open at the end of the test. So the fatigue sharpening has to be done by 'dead reckoning' - i.e., good judgment and experience.

Qu.: What are "side-grooves"?

CT specimens invariably employ side-grooves. Usually this involves 2.5mm deep grooves with a Charpy-like V-notch profile on both sides of the specimen. A 25mm thick CTS is thus locally reduced to a net thickness of $B_n = 20$mm, only 80% of the gross thickness.

Qu.: What is the purpose of side-grooves?

The benefits of side-grooves are,

- The crack is constrained to grow along the ligament $(\theta = 0)$ rather than veering off to one side, making the interpretation more difficult;
- The crack front of a non-side-grooved specimen tends to adopt a "thumb-nail" shape, due to the greater constraint in the middle of the section compared with the free surfaces. This again makes interpretation difficult since the crack growth, $\Delta a$, varies across the specimen rather than having a single, well defined, value. Side-grooves tend to increase constraint across the whole of the crack front, usually promoting a straighter crack front. (But sometimes they may overdo it and produce an inverse thumb nail).
- The side-grooves promote higher constraint, thus enhancing validity and ensuring conseravtism (i.e., growth rates will tend to be faster at the same C* than without a side-groove).

Qu.: What is the basic ccg testing methodology?

The methodology is broadly similar to that of low temperature fracture toughness J-testing, except, of course, that you must wait thousands of hours or more to acquire the data. There are three essential measurements,

- Load;
- Load line displacement - increasing over time;
- Crack length increment - increasing over time.

The usual procedure is to hold the load fixed throughout the ccg test and to monitor the load line displacement and crack growth over time.

Load line displacement is the strict requirement since this is conjugate to the load, i.e., $\int FdD$ provides an energy measure. However, options to measure displacement, say, at the crack mouth, exist. These are reliant on specimen-specific formulae for C* which implicitly correct for the difference.

Qu.: How is displacement measured?

Due to the high temperatures, resistance-based LVDTs are not preferred. Instead capacitance gauges of some form are usual. The design of these capacitance gauges differs markedly according to whether oxidation of the specimen during the test is likely (ferritics) or not (austenitics). Purpose-built rigs are used to calibrate each individual gauge before they are used in the ccg test.

Qu.: How is crack growth measured?

The usual method of monitoring crack growth during the progress of the test is to use DCPD (direct current potential drop). ACPD is not recommended in EDF Energy in the context of ccg testing (though some testing labs do use ACPD).

It is essential that the DCPD is calibrated for each test after the test has finished. This is done by breaking open the specimen and measuring the actual crack growth.

DCPD provides a measure of crack growth in real time during the test execution. This is essential since the interpretation of the test data involves plotting the $\dot{a} - C^*$ trajectory (see below). It is also necessary to judge when to stop the test. Why?

- If the crack growth is too small then the test fails to provide any useful crack growth data - and you only find out for sure that the crack growth was too small after breaking open the specimen, so the option for a re-start no longer exists;
- If the crack gets too large the growth rate will accelerate wildly and the specimen will fail. In this case calibration of the DCPD is no longer possible and the test cannot be interpreted correctly.

Consequently there is always a nice judgment in deciding when to terminate a ccg test.

Qu.: How does breaking the specimen provide the final crack growth?

Measurement of the creep crack growth relies upon being able to distinguish between the EDM notch, the fatigue grown region, the ccg and the fracture surface produced when the specimen is broken. The break-open is done at liquid nitrogen temperature, and the EDM and fatiguing are carried out at room temperature. Hence, the different mechanisms and temperatures compared with the ccg part of the fracture surface lead to a very different fracture appearance. Consequently there is usually no problem in making the distinction.

If the crack front is not straight (and generally it will not be exactly straight) then the total crack area is measured. The crack growth is then defined as the average, i.e., the area divided by the net thickness, $B_n$.

Qu.: How is the data interpreted?

Because the intention is derive a growth law in the form $\dot{a} = AC^{*q}$, where $\dot{a} \equiv \frac{da}{dt}$, the objective is to measure both $\dot{a}$ and $C^*$. The former is provided by the slope of the crack growth versus time curve obtained from DCPD. However $C^*$ is more indirect. It is estimated from the load and the measured displacement in a manner closely analogous to that in J-testing - except that it is displacement *rate* which is used in C*.

Thus we arrive at a sequence of points on the $\dot{a} - C^*$ plot as time increases. If the growth really does obey a law $\dot{a} = AC^{*q}$ then this trajectory will be linear on a log-log plot, with slope $q$.

Qu.: What is the basis of the C* estimation methodology?

To a first order approximation $C^*$ is estimated by,

$$C^* \approx \eta \frac{P\dot{\Delta}}{A} \tag{3}$$

where, $P$ is the (constant) load, $A$ is the net section area, $A = B_n(w-a)$, and $\dot{\Delta}$ is the displacement rate.

Qu.: How is the eta-factor $(\eta)$ in (3) found?

The eta factor $(\eta)$ depends upon crack size and specimen type, but not explicitly upon creep strain, stress or time. It will differ according to whether displacements are measured on the load line or the crack mouth. Expressions for the eta factors are given in Ref.[1]. They are roughly 1 for tension dominated specimens and roughly 2 for bending dominated specimens. The reason for this is the same as was discussed in the context of toughness testing in http://rickbradford.co.uk/T73S02TutorialNotes18.pdf.

Qu.: Whence comes (3) ?

The numerator in (3) is the rate of doing work on the specimen, so that (3) is essentially $C^* \approx \frac{dJ}{dt}$, where the $J$ in question is the $J(t)$ discussed in session 39 (which includes creep strains and hence differs from the low temperature J used in R6). We shall see that this J is essentially the "creep toughness".

Qu.: What is the full C* estimation procedure?

Equ.(3) is only a rough indication. The actual procedure recognises that C* should depend only upon that part of the displacement rate which results from the creep strains. Because the crack is growing the specimen is becoming more compliant. Consequently part of the increase in displacement is elastic (or elastic-plastic) rather than creep. If the elastic and plastic parts of the displacement are written $\Delta_e$ and $\Delta_p$ then the creep part is $\Delta_c = \Delta - \Delta_e - \Delta_p$. A formula like (3) then becomes correct so long as $\dot{\Delta}_c$ is used in place of $\dot{\Delta}$. Unfortunately the experiment only provides us directly with the total displacement, $\Delta$. The procedures of Refs.[1-4] require the elastic-plastic displacement rates to be calculated theoretical using formulae which they give, namely,

$$\dot{\Delta}_e = \frac{2(1-\nu^2)B_n K^2}{EP}\dot{a} \tag{4}$$

$$\dot{\Delta}_p = \frac{(1+n_p)B_n J_p}{P}\dot{a} \tag{5}$$

where $n_p$ is the plastic (Ramberg-Osgood) index and $J_p$ is the plastic part of $J$. Note that both these corrections to the displacement rate are proportional to the experimental crack growth rate.

Determination of $\dot{\Delta}_p$ might be expected to be problematical because it requires an estimate of $J_p$ which can only be derived from the test data and which requires identification of the plastic part of the energy (area under the load-displacement trace). Fortunately $\dot{\Delta}_p$ is most often small and can be neglected.

In contrast one might have expected $\dot{\Delta}_e$ to be rather accurately determined from Equ.(4) since the LEFM is accurately known simply in terms of the applied load and the current crack length. Even here, though, reality confounds us. There is a subtle but crucial difficulty that has only recently been appreciated (see below).

Qu.: Are there any further refinements to Equ.(3)?

Yes.

In addition to requiring the subtraction,

$$\dot{\Delta}_c = \dot{\Delta} - \dot{\Delta}_e - \dot{\Delta}_p \qquad (6)$$

There is an additional material-dependent factor, H,

$$C^* \approx H\eta \frac{P\dot{\Delta}_c}{A} \qquad (7)$$

Of course $H\eta$ could have been called just a single parameter. But the advantage of considering the two terms separately is that whilst $\eta$ depends upon crack length but does not depend upon the material, $H$ depends upon the material but not upon crack length. In fact, in most cases (but with some exceptions) $H = \frac{n}{n+1}$ for bending dominated specimens, and $H = \frac{n-1}{2(n+1)}$ for tension dominated specimens, where $n$ is the creep index.

Qu.: Is there a specimen size requirement

Recall that in J toughness testing there are certain requirements, such as a minimum size for the specimen as a multiple of $J/\sigma_y$ in order for the data to be valid. ASTM E1457, Ref.[2], does not specify an explicit size requirement in the same way that toughness testing standards do. Instead it requires that the time to achieve steady creep (C*) conditions be small compared with the test duration. This is ensured in EDF Energy practice (McLennon & Allport, Ref.[1]) via various criteria, listed below. However, note also that R5V4/5 Appendix A1, §A1.4.5.2 recommends explicit size requirements for a CTS in direct analogy to J-toughness testing, i.e., that the key dimensions ($a_0, w - a_0, B_n$) be greater than 25 times the 'initiation' CTOD (defined as the CTOD when crack growth reaches 0.2mm).

Qu.: What are the validity requirements?

The validity requirements from Ref.[1] are,

[1] Only data with crack extensions >0.2mm should be used to derive an $\dot{a} - C^*$ relation;

[2] Only data obtained after the transition time, $t_T$, should be used to derive an $\dot{a} - C^*$ relation. The transition time is given by,

$$t_T = \frac{(1-\nu^2)K^2}{(n+1)EC^*_{\exp,\min}} \qquad (8)$$

where $C^*_{\exp,\min}$ is the minimum experimentally determined value of $C^*$. In practice within EDF Energy the check under [6], which fulfils essentially the same purpose, is generally used instead of this. (The purpose in question is to ensure that data is used only in the C*, rather than the C(t), controlled regime).

[3] The creep displacement rate should exceed half the total displacement rate, $\dot{\Delta}_c > 0.5\dot{\Delta}$. In practice this restriction has commonly been ignored – more of this problem below.

[4] The load line displacement should not exceed 5% of the specimen width, w. This restriction may be unduly restrictive for creep ductile materials.

[5] Recall from http://rickbradford.co.uk/T73S03TutorialNotes39.pdf that the creeping HRR fields prevail at the crack tip only if the crack grows sufficiently slowly. Otherwise, for a moving crack tip, the fields will be of a different form, e.g., Hu-Riedel form. Consequently there is a requirement that the dimensionless crack velocity be less than 0.5, i.e.,

$$\lambda = \frac{\dot{a}\sigma_{ref}^2}{EC_{ref}^*} < 0.5 \quad (9)$$

[6] The final requirement is aimed at ensuring that transient creep has ceased so that $C^*$ rather than $C(t)$ is the controlling crack-tip field parameter. This is important so as to ensure that the empirical formulae discussed above [Equs.(3-7)] are valid. It also justifies the use of the symbol $C^*$ in Equs.(3,7). The criterion recommended in Ref.[1] is that data is valid after a redistribution time defined as when the reference creep strain reaches half the elastic reference strain, i.e., when $\varepsilon_c(\sigma_{ref}) = \sigma_{ref}/2E$.

Qu.: What is the effect of constraint on ccg rate?

There is an effect of constraint on ccg rates, as you would expect – more highly constrained conditions (plane strain) promoting faster growth, However I believe that the difference between growth rates in plane stress and plane strain is not as great as the factor of ~50 suggested by Webster & Ainsworth, Ref.[4]. This will be discussed in greater detail in http://rickbradford.co.uk/T73S03TutorialNotes42.pdf.

Indeed, the *experimental data* from Webster & Ainsworth (as opposed from theory) does not support such a very large constraint factor. This is illustrated for low alloy ferritic steels in Figures 1 - 3. Salient points are,

- The theoretical plane strain and plane stress lines drawn on Figures 1 - 3 will be derived in the next session (from continuum damage mechanics);
- The growth rates in Figures 1 and 2 increase as either the specimen becomes bigger or the side-grooves are made deeper, confirming the effect of constraint.
- The data in Figure 1 does support a large enhancement of growth rate due to constraint, though the largest and most highly constrained specimens suggest a factor of ~10 over plane stress growth rates (rather than ~50);
- The alarming thing is that specimens which (I guess) pass the size criteria and which one would therefore have expected to be fully constrained (e.g., a 2" x 2" x 1" CTS with 2.5mm side-grooves) actually produce results near the plane stress line, Figures 2, 3.

- Figure 2 is actually for deeply cracked bend (DCB) specimens, but these should be highly constrained (I think).
- The most highly constrained results (in Figure 1) were obtained from impractically huge specimens, some 12" x 10" x 2.5", which would have weighed ~85 lbs and required lifting equipment to handle.
- The only data in Webster & Ainsworth at the plane strain line was for an aluminium alloy.

Where does this leave us? It appears that the ideal of high constraint, which is the achievable goal in low temperature fracture, is simply not a practical possibility in creep crack growth – except for materials of sufficiently low creep ductility.

This is an issue which seems to have received little attention. The implicit article of faith would appear to be that, whilst your test specimens may have low constraint they are probably at least as constrained as the plant to which the data will be applied.

**Figure 1 Effect of Constraint on CCG (after Webster & Ainsworth)**

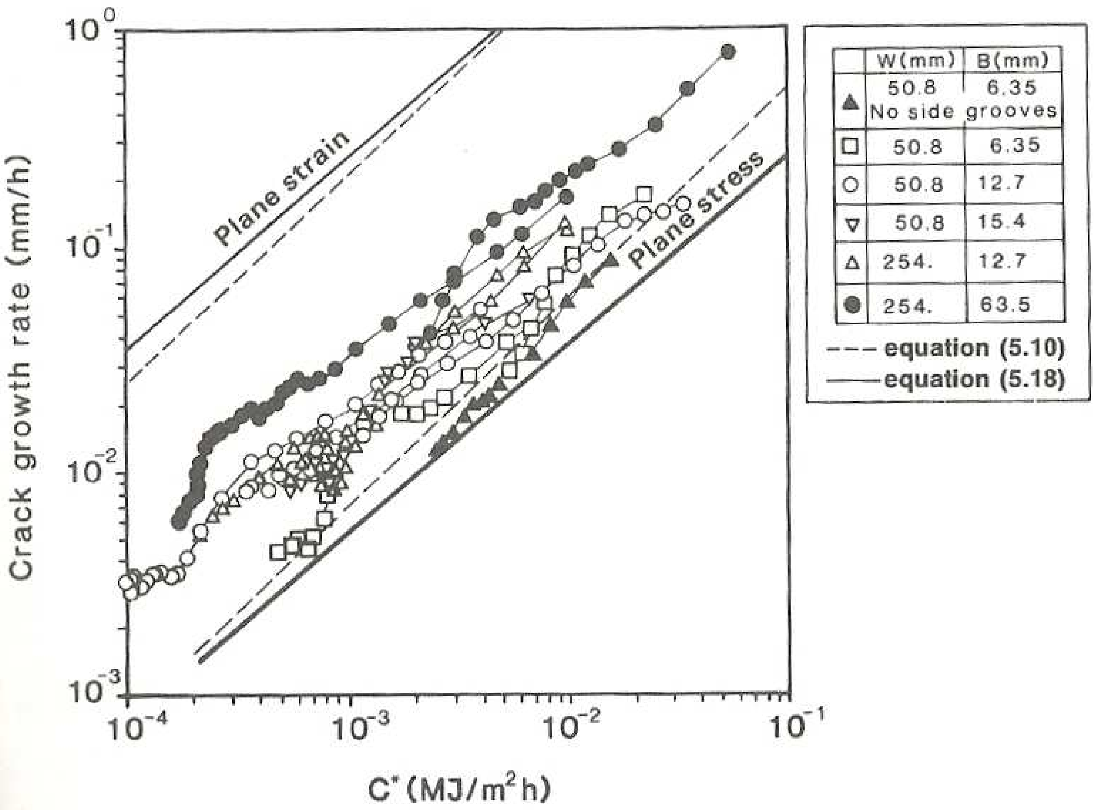



Fig. 5.11 Creep crack growth properties of different sizes of CT specimen of 1% CrMoV steel at 538 °C [19] with $(B_n/B) = 0.75$ compared with predictions.

**Figure 2** **Effect of Constraint on CCG (after Webster & Ainsworth)**

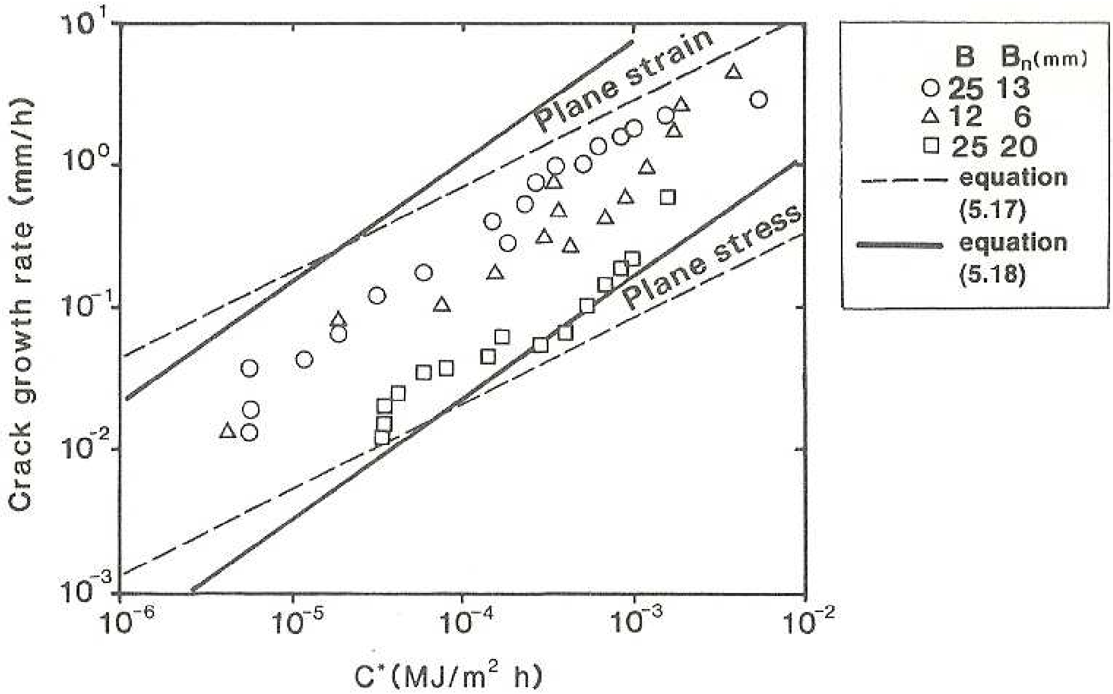



Fig. 5.14 Crack propagation characteristics of $\frac{1}{2}$% CrMoV steel obtained on DCB speci
at 565 °C [16].

**Figure 3** **Effect of Constraint on CCG (after Webster & Ainsworth)**
**NB: Plane Stress and Plane Strain labels should be reversed**

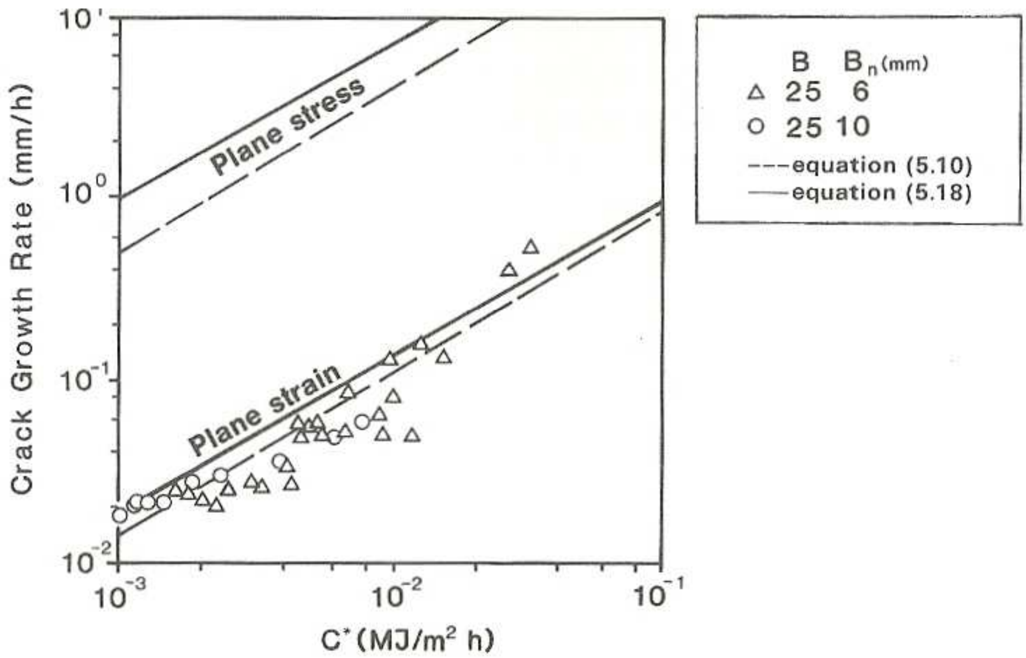



Fig. 5.9 Creep crack growth characteristics of $2\frac{1}{4}$% CrMo steel measured on CT specimens at 538 °C [18] compared with predictions.

**Qu.: Test versus Plant C* Magnitudes**

Figures 1 to 3 do not contain data below $10^{-6}$ MPa.m/hr, and this is typical of laboratory tests. Figure 4 shows some data for austenitic material (316 HAZ) which is also at C* values above $\sim 10^{-6}$ MPa.m/hr (except temporarily for one test reduced to 470$^{o}$C). This contrasts with plant where the typical C(t) or C* values are $\sim 10^{-8}$ MPa.m/hr, some two orders of magnitude smaller – at least in cases where the assessment implies acceptably little growth over plant life.

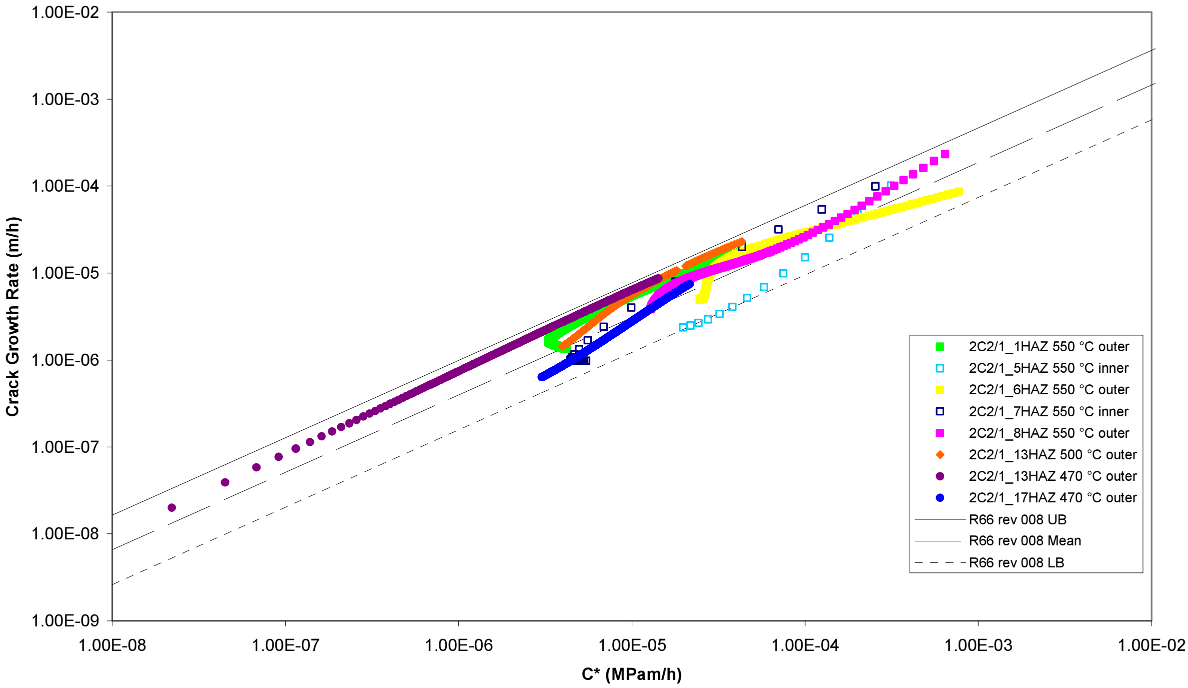
The reason is largely due to the difference in timescale of lab tests compared with Power station lives (40+ years if we're lucky). The tests in Figure 4 had durations in excess of two years in some cases, which is a good deal longer than would have been common up until about 12 years ago. And some of the series of 316H/316HAZ tests now being analysed have durations in excess of 5 years, albeit not many.

Indeed these recent longer term tests on 316 have achieved ~$10^{-7}$ MPa.m/hr. This latter value has begun to bring the experimentally determined range within that assessed for plant ($10^{-7}$ MPa.m/hr corresponding to an upper bound growth rate of ~1 mm/yr in 316, for example, which might be tolerable for a few years in thicker sections).

Consequently extrapolation of the experimental data is required for plant applications – and this generally assumes the straight line in log-log space implied by $\dot{a} = AC^{*q}$ is valid. The tendency has been that, as longer term data is acquired, this assumption is repeatedly challenged. Unfortunately, longer term tests, at lower stresses and lower strain rates, tend to exhibit lower creep ductility. This is manifest as tight cracks with little crack tip blunting or deformation. It can also be manifest as a tendency for the $\dot{a} - C^*$ line to drift upwards. Moreover, as a double jeopardy, the reduced ductility could lead to greater constraint, providing another reason for the $\dot{a} - C^*$ line to drift upwards.

Consequently the issue of extrapolation of ccg data to plant C* values is tricky and not satisfactory at present.

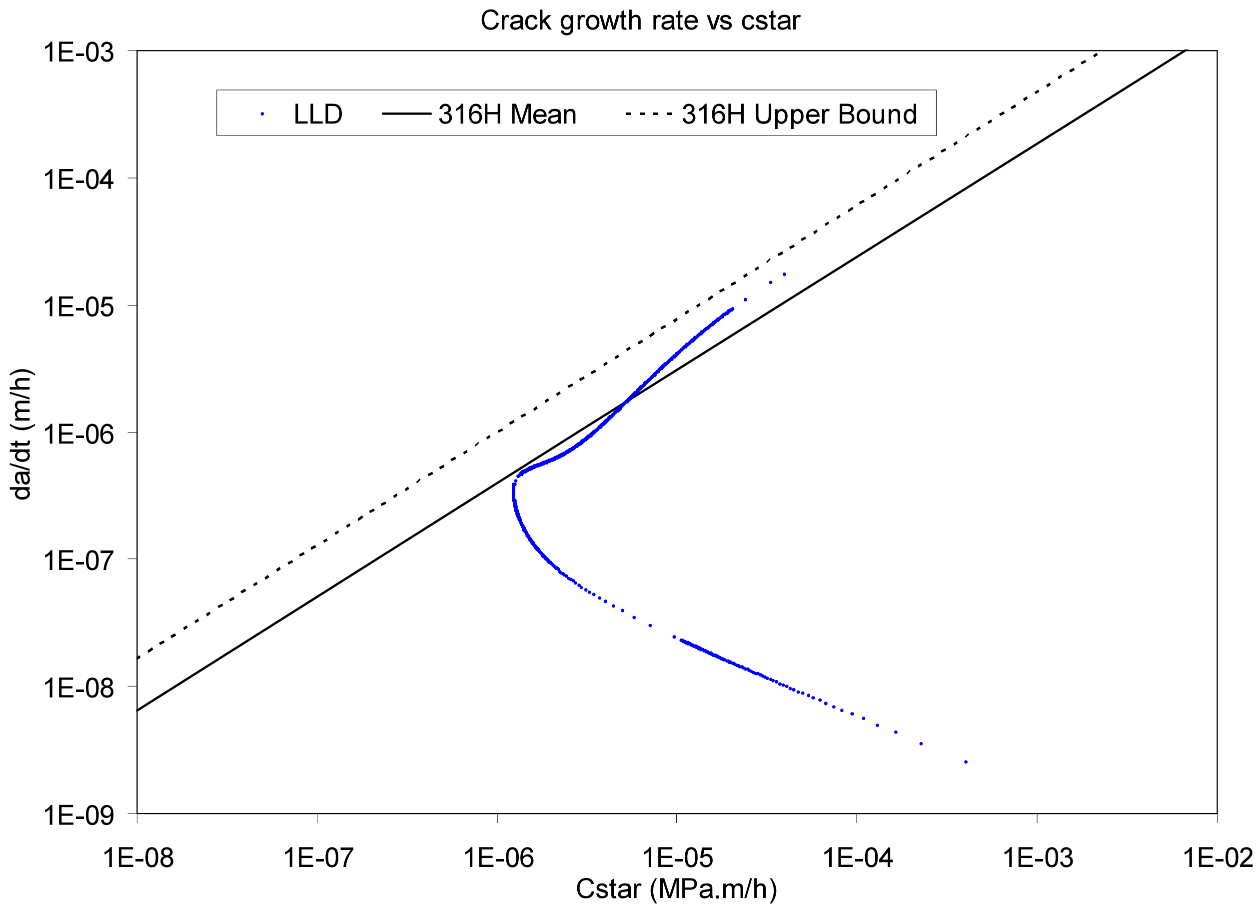
**Figure 4: da/dt – C* data for 316 HAZ ( from E/REP/BBGB/0035/GEN/08)**




Qu.: What are "tails"?

When the early data is plotted on the $\dot{a} - C^*$ plot it does not initially fall along the trend line shown in Figures 1 to 4. The characteristic behaviour on the trend line is that both $\dot{a}$ and C* are increasing – so the trajectory is moving upwards and to the right. But the initial data most often lies at low $\dot{a}$ but *high* apparent C*. This produces points which lie below the trend line and far to the right. As transient creep dies away, the $\dot{a} - C^*$ trajectory moves towards the trend line, C* decreasing. When the trend line is reached the trajectory changes direction to join the trend line, so that C* is now increasing. The initial trajectory off the trend line is known as the "tail". A typical tail is illustrated in Figure 5. However, tails are very variable and can even occur above the trend line, particularly if test conditions are changed. Figure 5 is most typical, though. [There is an argument that tails which lie above the trend line are indicative of C(t) control, as opposed to C* control, see http://rickbradford.co.uk/ATaleOfTwoTails.pdf ].

**Figure 5 Typical $\dot{a}-C^*$ tail (316 Parent, 525°C)**




Qu.: Why are there tails?

Recall that crack growth does not initiate immediately – there is an incubation period. During this period the creep damage around the crack tip accumulates until it reaches its steady state distribution corresponding to crack growth. Consequently $\dot{a}$ starts at zero (though the DCPD equipment may not give exactly zero). In contrast, C* is not zero since the creep strains, and hence the creep component of displacement, starts immediately. In fact, thanks to primary creep, the initial creep displacement rate is fast and decreases initially. Consequently one expects the starting point to be at very small $\dot{a}$ but large C*, i.e., a point at the bottom right of the graph. So there are three reasons for the tail,

- Incubation – the period whilst creep damage accumulates at the crack tip;
- Primary creep – initial strain rates reduce quickly;
- Redistribution of the high stresses at the crack tip, causing the initially C(t) dominated fields to give way to the C* dominated fields.

The validity requirements discussed above are intended to censor the tail data. In practice, however, rather than use the validity requirements the pragmatic approach is simply to see when the $\dot{a}-C^*$ trajectory turns around and becomes an upward sloping straight line, i.e., a trend line.

Qu.: Is $da/dt = A.C(t)^q$ justified?

The test procedure is designed to ensure that data is collected only in the *C** controlled regime, not the *C(t)* regime. Consequently test data produces a $da/dt = A.C^{*q}$ trend line, not a $da/dt = A.C(t)^q$ trend line.

In contrast, in applications to plant, at least for austenitic materials, it is generally $da/dt = A.C(t)^q$ which is used, since transient conditions often prevail for which *C(t)* is the correct parameter. Since $C(t) \geq C^*$ it is essential to use $C(t)$ to ensure conservatism.

However, I have never seen any experimental justification that $da/dt = A.C(t)^q$ gives a correct prescription for ccg when $C(t) > C^*$. The justification is entirely theoretical, i.e., on the grounds that $C(t)$ controls the crack tip fields.

For ferritic steels this may not matter much because the C* regime tends to be achieved quickly. However, for austenitics the C(t) regime may prevail for a long time, particularly if secondary stresses are large - perhaps for the whole plant life.

Qu.: Can C(t) be measured?

I am not aware of any empirical estimation procedures for C(t). I suspect it may not be feasible. This is because when C(t) is substantially larger than C* it is likely that the reference creep strain is too small to be discernable from displacement measurements. The "C(t) effect" is essentially confined to near the crack tip and probably cannot be measured by gross quantities controlled by reference stress or reference strains.

Qu.: What is the problem with 316 HAZ?

The problem with thick section 316 HAZ that came to light around 2009 is simply that, for the same applied reference stress, creep crack growth rates through HAZ appear to be far faster than through the parent material. This is illustrated in Figure 6. The difference is large – significantly more than a factor of 10.

A very similar difference is seen in the displacement rates for HAZ specimens compared with those for parent specimens at the same applied load (Figure 7).

This is embarrassing because the current ccg law used for 316 HAZ in plant applications is essentially the same as the parent law.

Why has this rather emphatic difference only been discovered so recently? The reason is that when plotted in $\dot{a} - C^*$ form the parent and HAZ data follow the same trend line (and this is how previous test data has always been presented).

That this is so can readily be seen. Since the empirical C* estimation formula is proportional to displacement rate, we can use displacement rate as a surrogate for C*. Figure 8 shows crack growth rate plotted against displacement rate. The parent and HAZ a are seen to over-plot quite well. (In fact the lower temperature data which has been included in Figure 8 also over-plots well).

Qu.: What does this mean for plant assessments?

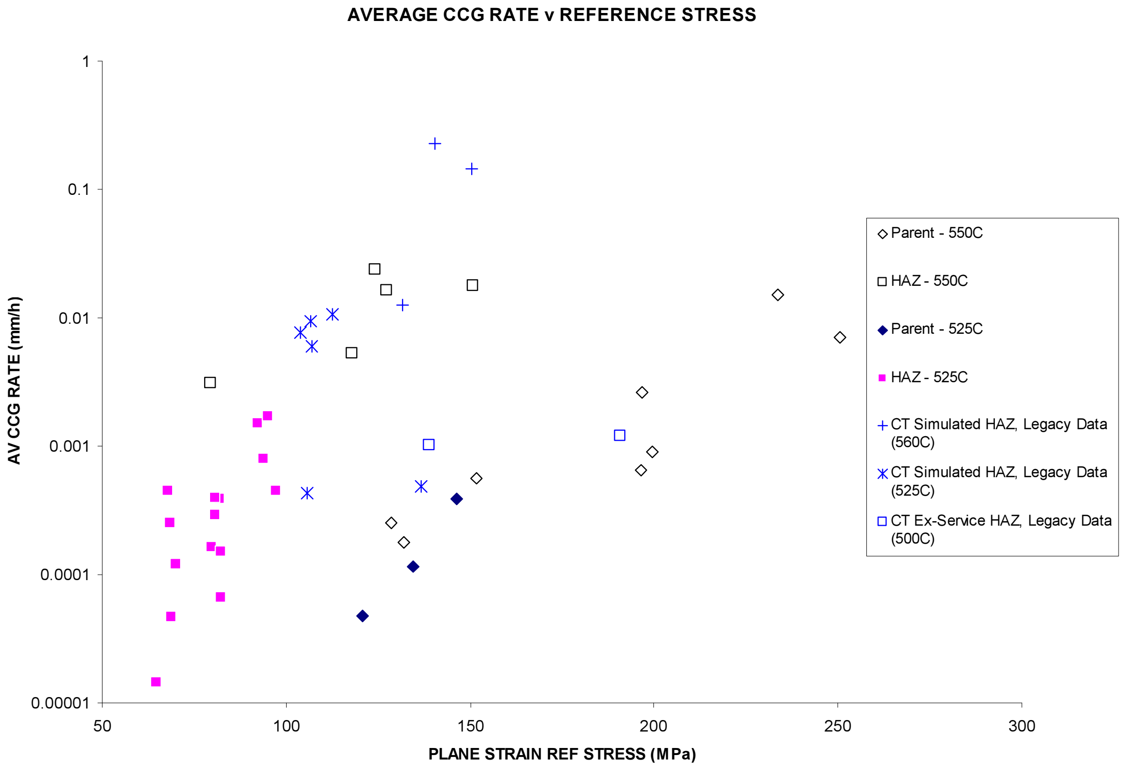
<u>This bit is in need of an update - Louise?</u>

One interpretation is that we have got the $da/dt = A.C^{*q}$ growth law right, but that the deformation rate of HAZ is far faster than we normally assume (i.e., RCC-MR parent). But this is probably not the right explanation. The evidence from HAZ or cross-weld creep tests is that HAZ creeps more slowly than parent, if anything (Ref.[6]).

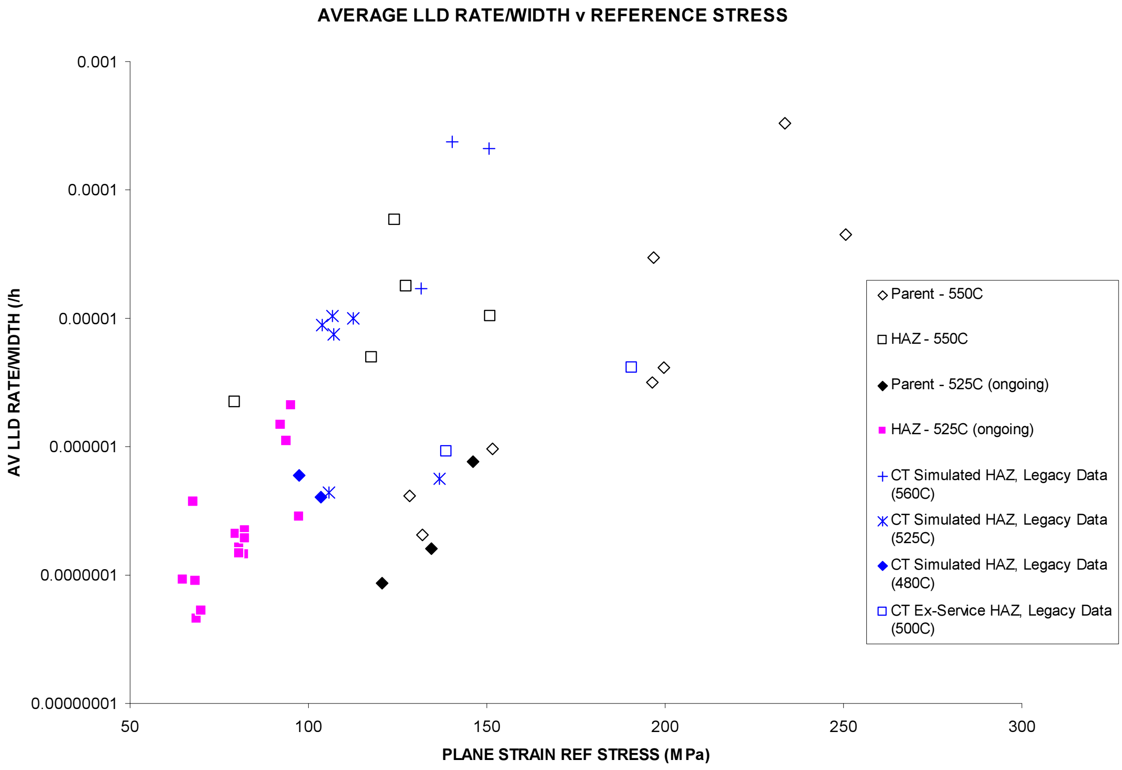
For at least some series of test data the explanation is likely to be the influence of welding residual stresses. These are known to be present in the specimens whose data is plotted in Figures 6-8. Consequently the effective stress experienced by the HAZ specimens is larger than indicated in Figures 6 and 7 – which could potentially bring the HAZ data into alignment with the parent data.

However, not all tests series have yet been quantitatively explained in terms of residual stresses. It may be significant that residual stresses are triaxial in nature, and this may affect how they relax.

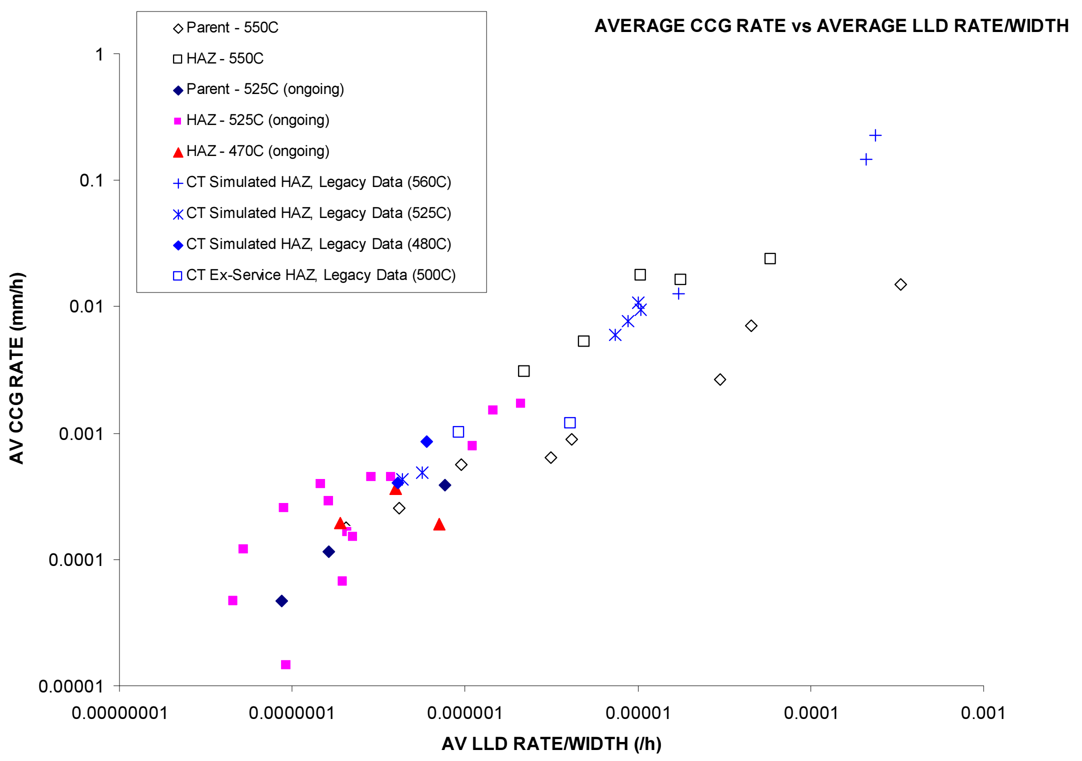
**Figure 6 316 HAZ cf Parent Crack Growth Rates (at same applied stress)**



## Figure 7 316 HAZ cf Parent CTS Displacement Rates (same applied stress)




## Figure 8 316 HAZ and Parent Crack Growth Rate v Displacement Rate



Qu.: What is the "subtraction problem"?

This is another bit which could do with checking against current understanding

This is another problem which came to light initially in the context of 316 HAZ, but which is really a more general problem. Recall from Equs.(6,7) that the elastic-plastic displacement rate should be subtracted from the total displacement rate in order to find the creep displacement rate, $\dot{\Delta}_c = \dot{\Delta} - \dot{\Delta}_e - \dot{\Delta}_p$, from which C* is calculated. The problem comes when the estimated elastic displacement rate, $\dot{\Delta}_e$, found from equ.(4), is virtually the same as, or even bigger than, the measured total displacement rate, $\dot{\Delta}$. Even for zero plastic rate, this leaves a creep displacement rate, $\dot{\Delta}_c = \dot{\Delta} - \dot{\Delta}_e$, which is virtually zero, or even negative.

Potentially this could undermine the whole process of experimental C* estimation, and hence the empirical growth law. In practice, to date, the problem has been avoided (rather than solved) by the simple, but unjustified, expedient of ignoring the subtraction and simple using the total displacement rate as if it were the creep displacement rate. This situation is unsatisfactory since the resulting growth law would appear to be potentially non-conservative.

A possible explanation for the subtraction problem is that creep crack growth consists of disconnected micro-cracking, linked by small islands of uncracked ligament which 'bridge' the crack. Ideally, measurements of the unloading compliance at the end of a ccg test would give an indication of the effective $\Delta_e$ for subtraction purpose. For example, if the unloading compliance at the end of the test is the same as that at the start, then no subtraction would be justified. Unfortunately, many tests did not carry out the (frustrating simple) procedure of unloading compliance measurement at the end of the test - and this cannot be done in retrospect if the specimens have been broken open.

A more disconcerting explanation is that it is indeed true that the bulk of the observed displacement is simple elastic opening of the specimen due to crack growth - with very little creep deformation actually taking place. Clearly creep damage is occurring at the cracks tips, because the crack is extending. But if creep damage and creep deformation are distinct mechanisms, as I believe, then it is perfectly possible for crack growth to occur when creep deformation is unmeasurably small.

## References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