Tutorial Session 31 (T73S04: R5V2/3) - Materials Data Requirements

Last Update: 2/2/15

Relevant to Knowledge & Skills items 1.2, 1.3, 1.4, 1.5

Materials data required; Ramberg-Osgood: differences between monotonic and cyclic; Conservative estimation of dwell stress; Continuous cycling fatigue data versus endurance with-dwells; Creep ductility: dependence on strain rate and triaxiality; Stress dependent ductility formulation; Spindler fraction; thermal ageing effects; Elastic follow-up factor (Z); Definition of creep damage; Reheat cracking in brief: how Z, S and ε_f produce cracking;

Qu.: What materials data are required for an R5V2/3 assessment?

A lot!

The required material data are listed in R5V2/3 Section 5 with further details in Appendix A1. In summary they are,

[1] Elastic moduli, E, ν;

[2] The 0.2% proof strength (and possibly the whole monotonic stress-strain curve, but not in general);

[3] A cyclic stress-strain curve (of which more below);

[4] Fatigue endurance (i.e., the S-N curve – for continuous cycling);

[5] Creep rupture data/equation;

[6] Creep deformation equation (or possibly a relaxation equation instead), plus the creep index, n;

[7] Creep ductility, including its dependence on strain rate and stress triaxiality, and possibly dwell stress.

[8] Ks – The shakedown factor, which is sometimes thought of as being an 'assessment parameter', but it is actually materials data, derived from materials tests.

Notes:

(i) An alternative to [5], [6] and [7] if creep is insignificant is an insignificant creep curve. Otherwise insignificant creep can be demonstrated from a deformation equation which extends to sufficiently low temperature, sufficiently long times and sufficiently high stress, to evaluate the insignificant creep condition.

(ii) An alternative to a creep deformation equation would be a stress-relaxation fit, but even in this case the creep deformation exponent, n, is required as well (since it appears explicitly in several places in the R5 procedure).

(iii) Although not explicitly listed above, if there are thermal stresses then the coefficient of thermal expansion, α, is required. If it is necessary to solve a thermal transient problem to find the temperature distribution leading to the thermal stressing, then further materials data will enter the problem (e.g., the thermal conductivity, the specific heat, and the density of the steel, plus data relevant to the fluid/boundary conditions).

Qu.: What does "monotonic tensile data" mean?

The monotonic tensile data is simply the result of doing a standard tensile test with increasing, non-reversing, load. This produces the standard definition of 0.2% proof strength, 1% proof strength and UTS. The 0.2% proof strength will certainly be required in the R5V2/3 procedure.

The whole of the monotonic stress-strain curve *may* be required also. This is generally specified in Ramberg-Osgood form. This expresses the elastic-plastic strain as the sum of the elastic strain and the plastic strain, with the latter being given by a power law in stress. Hence,

$$\varepsilon_{ep} = \frac{\sigma}{E} + \left(\frac{\sigma}{A'}\right)^{m'} \tag{1}$$

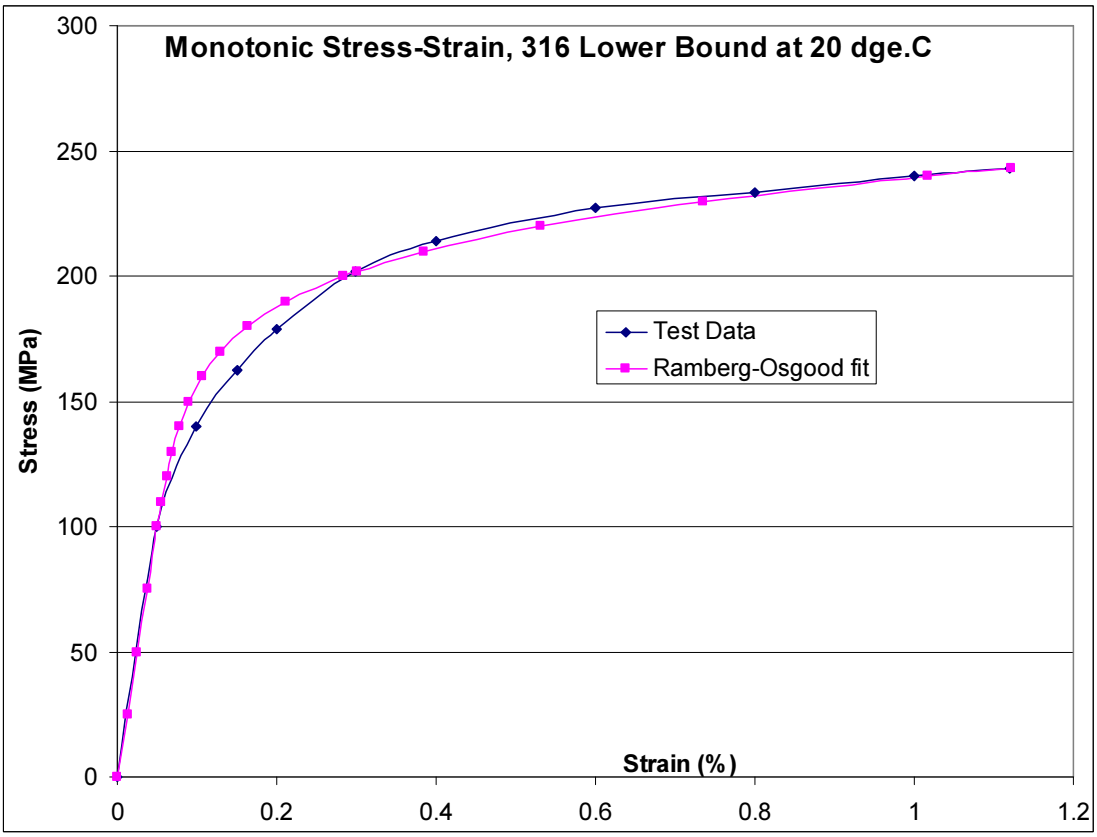
The parameter A' in (1) is generally written with a dash to distinguish it from the corresponding parameter for cyclic stress and strain (below). In R66 Section 3.1.1, A' is written as K', but it's the same parameter. Its units are MPa.

The plastic index, m', is often written as a reciprocal: $m' = 1/\beta'$, e.g., R5V2/3 Appendix A1, Equ.(A1.1), where the dash again distinguishes this parameter from the equivalent for cyclic stress and strain. (R66 Section 3.1 writes it as $1/n$, but this is to be avoided due to confusion with the creep index, n).

We expect $m' > 1$, typically between 3 and 10.

An illustration is shown in Figure 1 for 316ss parent at 20°C (lower bound), for which the Ramberg-Osgood fit is $A' = 412$ MPa, $m' = 8.71$.

Figure 1



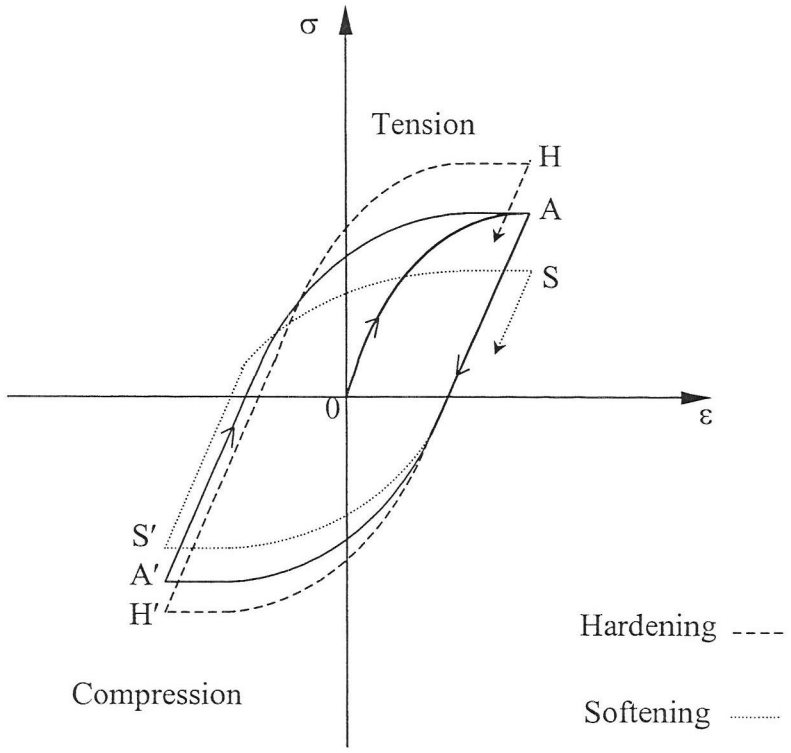
Qu.: What is the "cyclic" stress-strain curve?

When a material is taken around a load cycle involving yielding in both the tensile and compressive senses, the stress-strain cycles do not immediately over-plot. It takes a number of cycles to converge to a stable cyclic state so that the hysteresis loops over-plot. When they do, the shape of these loops is not the same as the monotonic stress-strain curve – as illustrated by Figures 2a and 2b.

Qu.: What is cyclic hardening and cyclic softening?

During the transient period before the steady cyclic state is attained, the hysteresis loops may either move to higher stresses or lower stresses, as illustrated in Figure 2. This is cyclic hardening or cyclic softening, illustrated by loops H or S in Figure 2 respectively.

Figure 2a



In truth, materials may cyclically harden over the first period and then cyclically soften after a certain number of cycles – see Figure 2c – or vice-versa! And this is for the ideal case when all load cycles are identical in terms of the applied strain, and the cycling occurs at constant temperature. In contrast, plant cycles are all different and generally involve large temperature changes. So it should always be born in mind that our assessments are only rough approximations to reality.

Figure 2b Creep-Fatigue Test, 316H at 550°C: Stress-Strain Hysteresis Loops

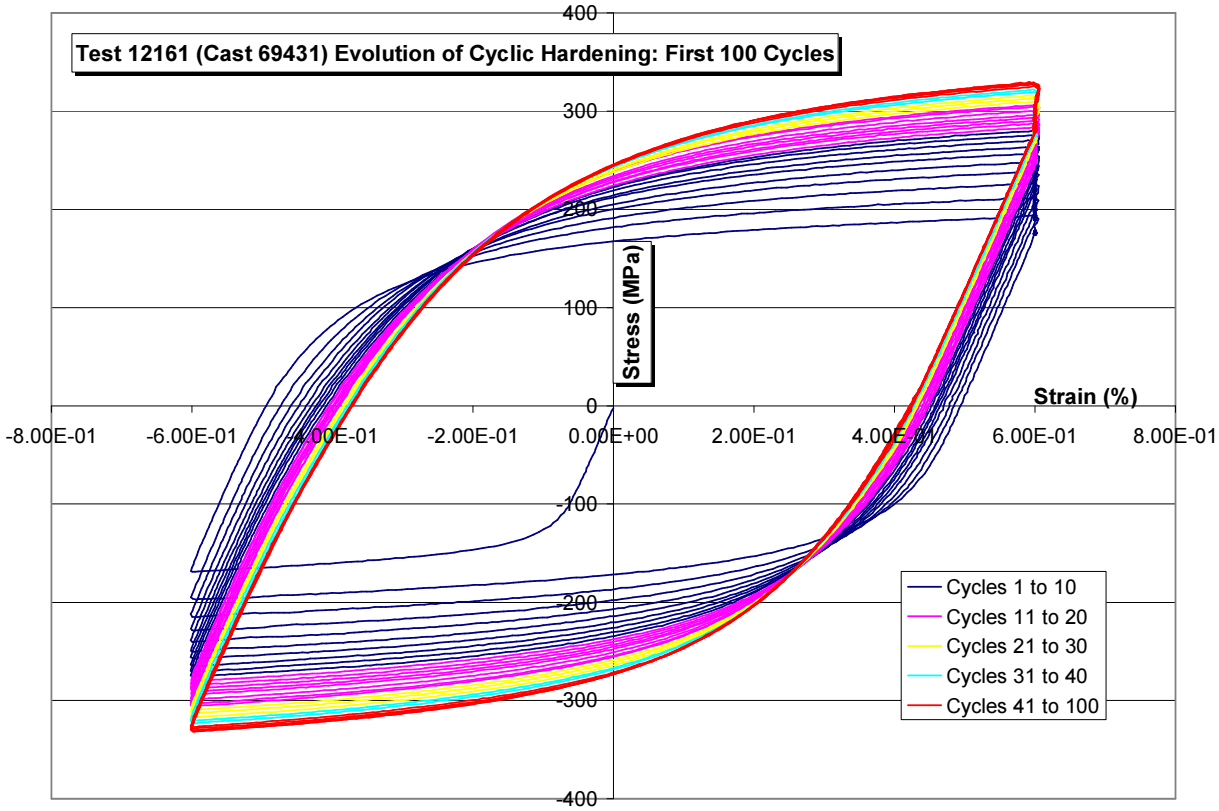
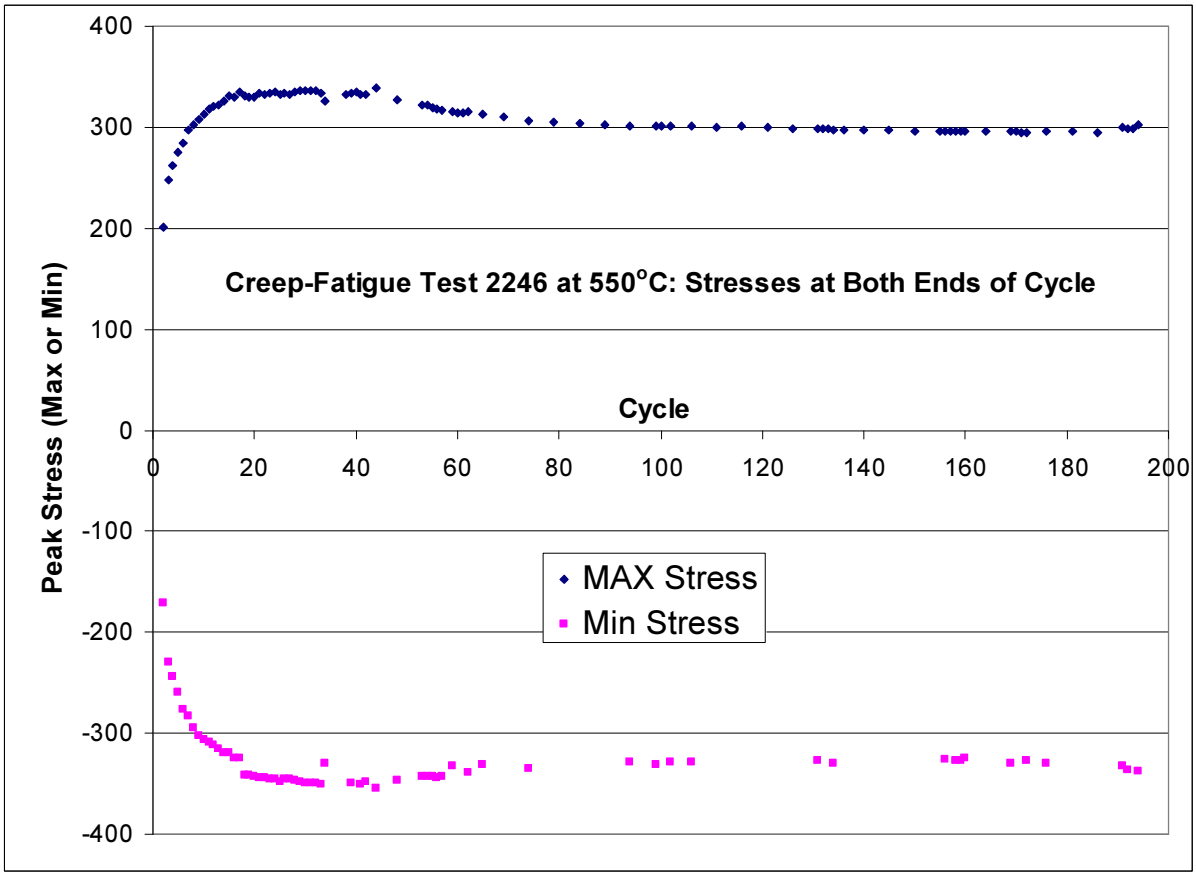


Figure 2c Creep-Fatigue Test, 316H at 550°C: Cyclic Max and Min Stresses



Qu.: Is there just one type of cyclic stress-strain curve?

No.

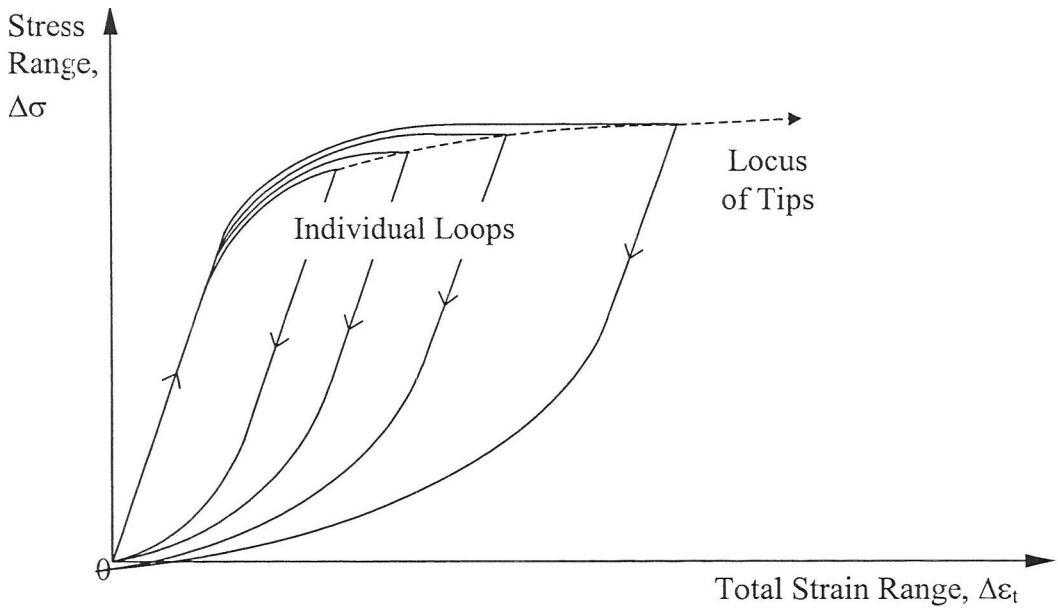
You may wish to know the relationship between the stress range and the strain range for the steady cyclic state. This defines the locus of the hysteresis loop tips, not the shape of the loops themselves, as illustrated by Figure 3.

This is what is conventionally meant by the "cyclic stress-strain curve". It is what is given in R66 Section 8. Such cyclic stress-strain curves are generally expressed in the Ramberg-Osgood form,

$$\Delta\varepsilon_{ep} = \frac{\Delta\sigma}{\overline{E}} + \left(\frac{\Delta\sigma}{A}\right)^{1/\beta} \tag{2}$$

where the parameters A, β will generally be different from the monotonic values.

Figure 3



Qu.: What's the other sort?

Alternatively, you may wish to know the shape of the loops themselves. With respect to an origin shifted to where the upwards going (left-hand) curve crosses the strain axis, as illustrated by Figure 4, an approximate expression for the curve shape is,

For $\sigma < 0$:

$$\varepsilon_{ep} = \frac{\sigma}{\overline{E}} \tag{3a}$$

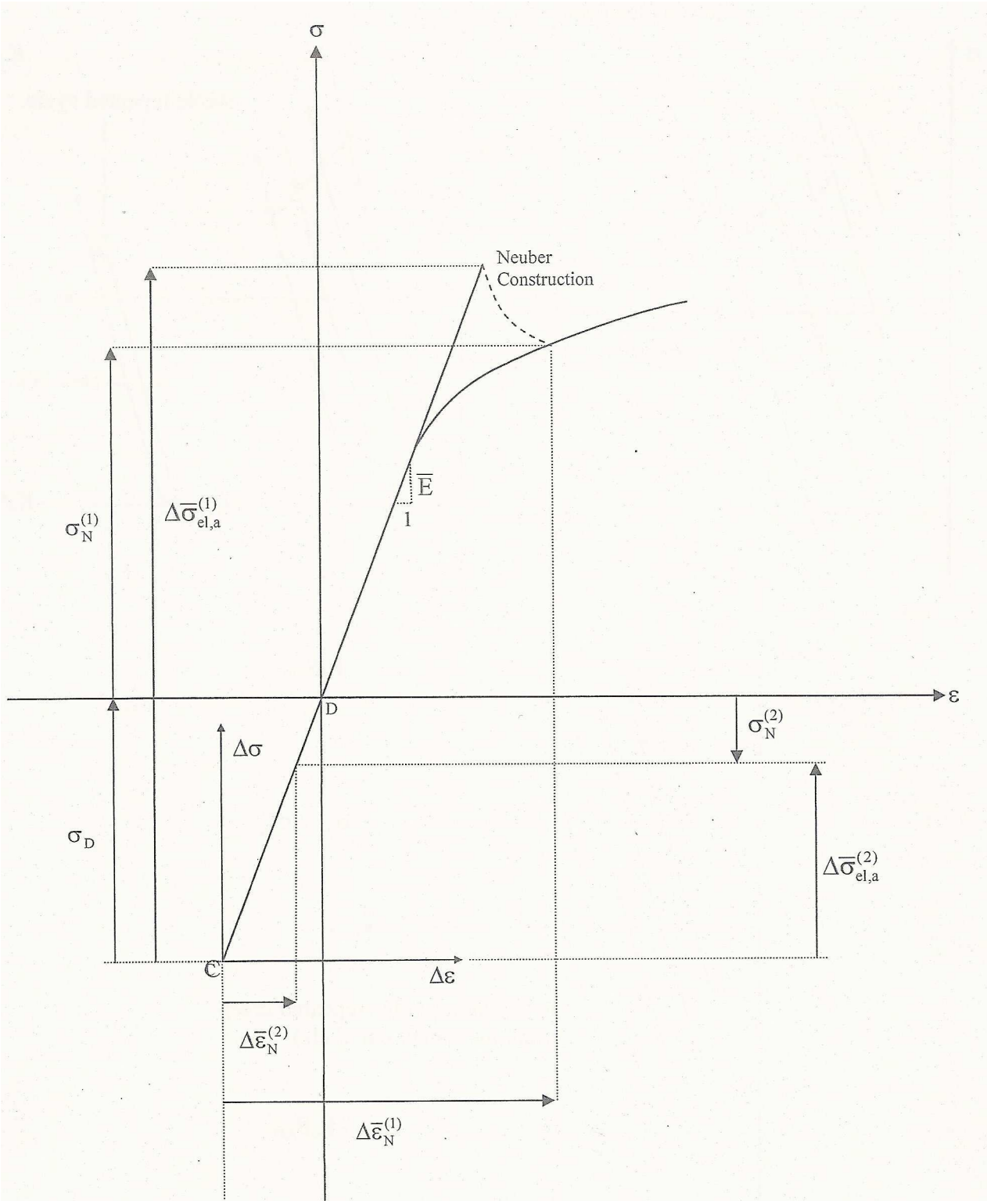
For $\sigma > 0$:

$$\varepsilon_{ep} = \frac{\sigma}{\overline{E}} + \left(\frac{2\sigma}{A}\right)^{1/\beta} \tag{3b}$$

Note that the parameters $\overline{E}, A, \beta$ appearing in (3b) are those derived from the cyclic stress-strain curve, (2), not the monotonic values. But (3b) differs from (2) due to the factor of 2 multiplying the stress in the plastic term. Note that it is the absolute stress which appears in (3a,b), not the stress range. And recall the false origin for the strain.

So the curve defined by (3a,b) is different from both the cyclic curve, (2), and the monotonic curve, (1). I shall refer to it as the 'modified curve'.

Figure 4: Illustrating the Origin for the 'Modified' Stress-Strain Curve



Qu.: So does R5V2/3 use (1), (2) or (3)?

Yes.

Qu.: What, that wasn't too helpful?

We will go into the gory details in session 33. For now note only that both (2) and (3) are used in the hysteresis cycle construction of R5V2/3 Appendix A7. Specifically,

- The 'standard' cyclic stress-strain curve, (2), is used for the side of the hysteresis loop without the creep dwell. This is because this side of the loop covers the full stress range, $\Delta\sigma$, and we only need to find the corresponding total $\Delta\varepsilon_{ep}$. The shape of the curve is then irrelevant so long as we get the correct tip-to-tip relation between $\Delta\sigma$ and $\Delta\varepsilon_{ep}$, which is provided by (2);

- But the other side of the hysteresis loop is interrupted by the creep dwell. In general this splits the plastic curve into two parts, neither of which spans the full stress range. Consequently it is necessary to use an equation describing the shape of the curve itself in order to relate the absolute stress, σ, to the increment of elastic-plastic strain. Hence (3) is the relevant equation.

Qu.: Does (1) get used too?

Typically you will not need the monotonic stress-strain curve, (1), in an R5V2/3 assessment (although the associated 0.2% proof stress is required, of course). However, in principle it can play a part because an effective cyclic curve can be constructed from the monotonic curve – see R5V2/3 Appendix A7, Section A7.2.2(2). [Note that the "zero hour isochronous" stress-strain curve is the same thing as the monotonic curve]. However this is unlikely to be a valid approach for typical BE assessments. There are a number of reasons why, but one is that the construction of Section A7.2.2 is not conservative when creep is the dominant damage mechanism.

Qu.: How are the Ramberg-Osgood curves used for a general 3D stress state?

R5V2/3 assessments are based for the most part on Mises equivalent quantities (there is an exception in some cases for the fatigue damage component, see Session 35). So the stress and strain, or stress and strain ranges, in (1), (2), (3) are to be interpreted as equivalent Mises quantities. In the case of *ranges*, note that R5 defines the equivalent quantity by forming the component ranges first and then finding the Mises equivalent from the component ranges.

Qu.: In (2) and (3) is E the ordinary Young's modulus?

No. The term $\overline{E}$ denotes the modified modulus.

For the reasons discussed in R5V2/3 Appendix A7, Section A7.2.1, E should be interpreted as taking the modified value,

$$E \rightarrow \overline{E} = \frac{3E}{2(1+\nu)} \tag{4}$$

The reasons for this are also discussed in Session 32.

Qu.: Should S_y be the lower bound or the mean 0.2% proof stress?

R5 defines S_y to be the lower bound 0.2% proof stress.

Qu.: Is the shakedown factor, K_S, used in the procedure apart from in the shakedown assessment?

Yes.

Recall that $2K_S$ is the factor by which the stress range must exceed the 0.2% proof stress (S_y) before ratcheting occurs, and so it is clearly important in the assessment of shakedown.

But K_S is also used in determining the absolute position of the hysteresis cycle in the vertical (stress) direction. Consequently it affects the dwell stress and hence the creep damage – which is usually the dominant damage term. So the damage can be sensitive to the assumed value of K_S. This is described in detail in the next couple of sessions.

Qu.: Should the cyclic stress-strain curves, (2) and (3), be mean or lower bound?

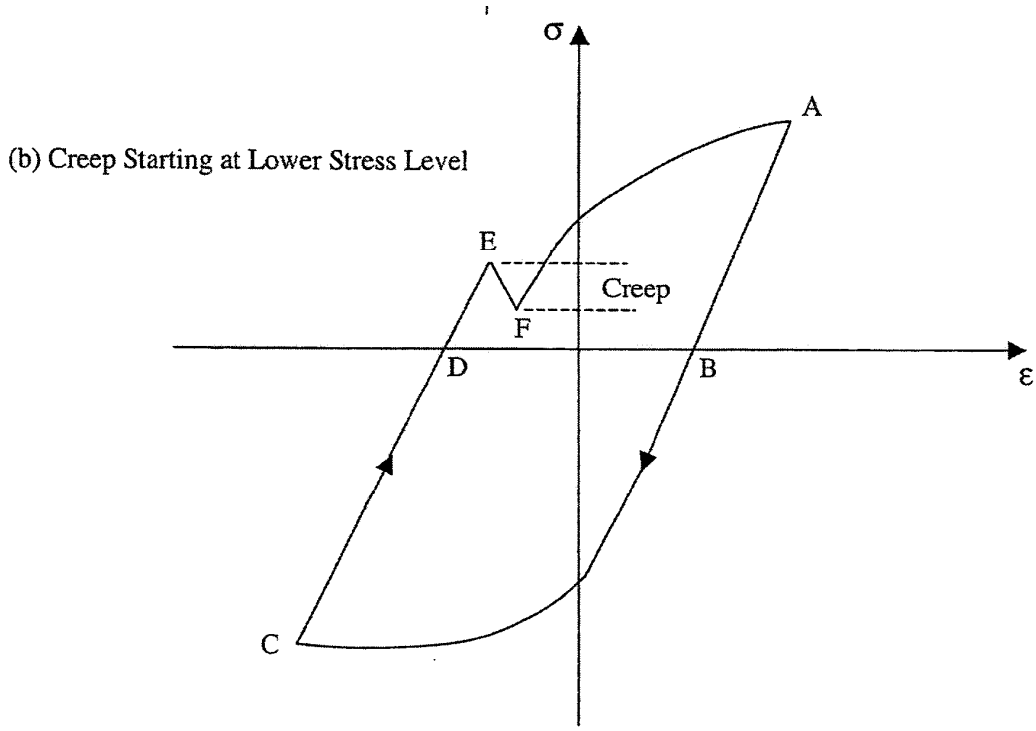
R5V2/3 Appendix A7 was originally designed to produce a conservative overestimate of the strain range, and hence to be conservative for fatigue (see Section A7.1). This is achieved by using a curve which under-estimates the stress range for a given strain range (and hence overestimates the strain range for a given stress range).

But our assessments are generally creep dominated. This requires a dwell stress which errs on the high side in order to be conservative. Consequently, if the hysteresis loop construction of Appendix A7 is used to estimate the dwell stress (and this is the best method, in my opinion) then which bound of the cyclic stress-strain curve is more conservative varies with circumstance, as follows.

Case A

If the dwell stress is defined via a stress increase from some reverse stress datum, such as is illustrated in Figure 5, then using a lower bound cyclic stress-strain curve will be conservative. This is because, for a fixed stress range between C and E, the absolute stress at E is greater the *smaller* is the magnitude of the compressive stress at C.

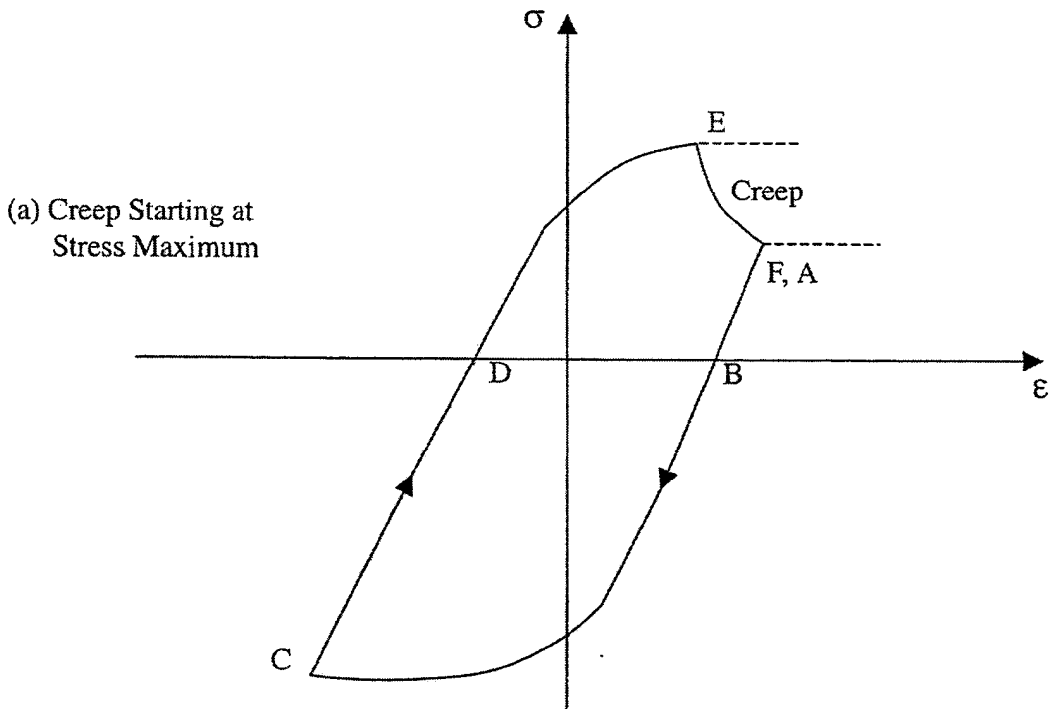
Figure 5



Case B

On the other hand, if the creep dwell occurs at the peak of the loop (as illustrated by Figure 6), or if the dwell stress is defined by a fixed stress drop from a forward stress datum, then it is conservative to use a high estimate for the cyclic stress-strain curve.

Figure 6



This is because, for a dwell at the peak of the cycle, the dwell stress is essentially the same as the cyclic peak stress. For a dwell below the peak, on the right, the dwell stress is the cyclic peak stress minus the stress drop between the peak and dwell. In both cases the dwell stress is larger *the higher* is the cyclic stress-strain curve.

Qu.: So which bound should we use for the cyclic stress-strain curve?

My advice is to use the best estimate curve in either case A or B, initially.

The reason is that the general philosophy in structural assessments, when estimating stresses, is to use a best estimate approach. Conservatism is ensured by using bounding properties for the 'strength' type materials data (e.g., endurance, ductility, rupture strength, etc.).

But there is potentially considerable sensitivity to this assumption and this should be explored by repeating the assessments with other reasonable variants. Whether the lower bound or the upper bound cyclic stress-strain curve is likely to increase the damage estimate can be gauged from the above observations.

Qu.: How important is creep ductility?

Creep ductility, ε_f, is often the most important piece of data in an R5V2/3 assessment. In the high temperature regions of AGRs such assessments are usually creep dominated, and the creep damage is, roughly,

$$D_c \approx \frac{\varepsilon_c}{\varepsilon_f} \qquad \text{(roughly)}$$

Consequently, whilst a great deal of calculational effort may have gone into estimating the creep strain, ε_c, the life of the component is still proportional to the assumed value for the creep ductility, ε_f. Unfortunately, the creep ductility is generally poorly known (or, to be more accurate, large quantities of test data display very large scatter). This seriously limits the accuracy of creep-fatigue initiation assessments to R5V2/3. It actually motivates the use of probabilistics, but I'll not get on my hobby horse just now.

Qu.: Is the creep ductility different from the plastic ductility?

Yes.

They are completely different quantities.

Qu.: How is creep ductility defined?

Opinions differ, but one definition is to take the uniaxial creep ductility as being the ***creep strain*** at failure in a creep test (or, alternatively, the last logged creep strain). The main alternative is to use the reduction of area in the failed region, which is invariably larger. But this is not the usual definition.

The important thing here is that the elastic-plastic strain must be subtracted from the total strain to leave only the creep strain before defining the creep ductility. Since the plastic strain may be substantially larger than the creep strain, this is extremely important.

Until around the mid1990s, R66 still contained data for some austenitics in which the plastic loading strain had been left unsubtracted – thus giving an erroneously high apparent value for the creep ductility. Historically this created a difficulty initially in

understanding austenitic reheat cracking – until this interpretation of the data was clarified.

Note that Mike Spindler's stress modified ductility approach treats the plastic component of the strain rather differently (see below).

Qu.: How is the multiaxial creep ductility defined?

In multiaxial conditions, the creep ductility is defined as the Mises equivalent creep strain (formed from the individual components of creep strain),

$$\varepsilon_{ij}^{c} = \varepsilon_{ij}^{total} - \varepsilon_{ij}^{ep}$$

$$\bar{\varepsilon}^{c} = \frac{\sqrt{2}}{3}\left\{\left(\varepsilon_{11}^{c} - \varepsilon_{22}^{c}\right)^2 + \left(\varepsilon_{22}^{c} - \varepsilon_{33}^{c}\right)^2 + \left(\varepsilon_{33}^{c} - \varepsilon_{11}^{c}\right)^2 + 6\left(\left(\varepsilon_{12}^{c}\right)^2 + \left(\varepsilon_{23}^{c}\right)^2 + \left(\varepsilon_{31}^{c}\right)^2\right)\right\}$$

Qu.: How is the multiaxial creep ductility found experimentally?

Well, obviously by using tests under multiaxial conditions. Ideally this would involve specimens stresses homogeneously. But this is hard to achieve under triaxial stress conditions. Homogeneous tests under biaxial, of nearly biaxial, conditions are practical, e.g., tension-torsion tests or pressurised cylinders with extra end-loads. However, test with severe triaxially are in practice inhomogeneous. Notched bars are most often used to determine multiaxial ductility (or to validate models of multiaxial ductility). But these tests involve only very localised triaxiality and steep stress gradients and so are not ideal.

Qu.: Does creep ductility vary much between weld, HAZ and parent materials?

Yes.

Generally ductility varies a great deal between the different weldment zones.

For example, for 316H parent welded with a typical 316 consumable, the weld ductility is usually greater than that of the HAZ (and probably greater than the parent also). For this reason reheat cracking in 316 weld material is (almost) unknown.

But austenitic weld material is not always superior to parent. For example, 347ss weld material often has very poor creep ductility and does reheat crack.

Qu.: How does creep ductility vary with strain rate?

This has already been discussed in Session 23. Figure 6 is a reminder for 316H. At slow strain rates the creep ductility can be far smaller (lower shelf). This means that accelerating creep tests can give a misleading, non-conservative, result for creep ductility.

Fig.7a Ex-HYA S/H Header 316H 525-575°C Creep Ductility (R66 Rev.008)

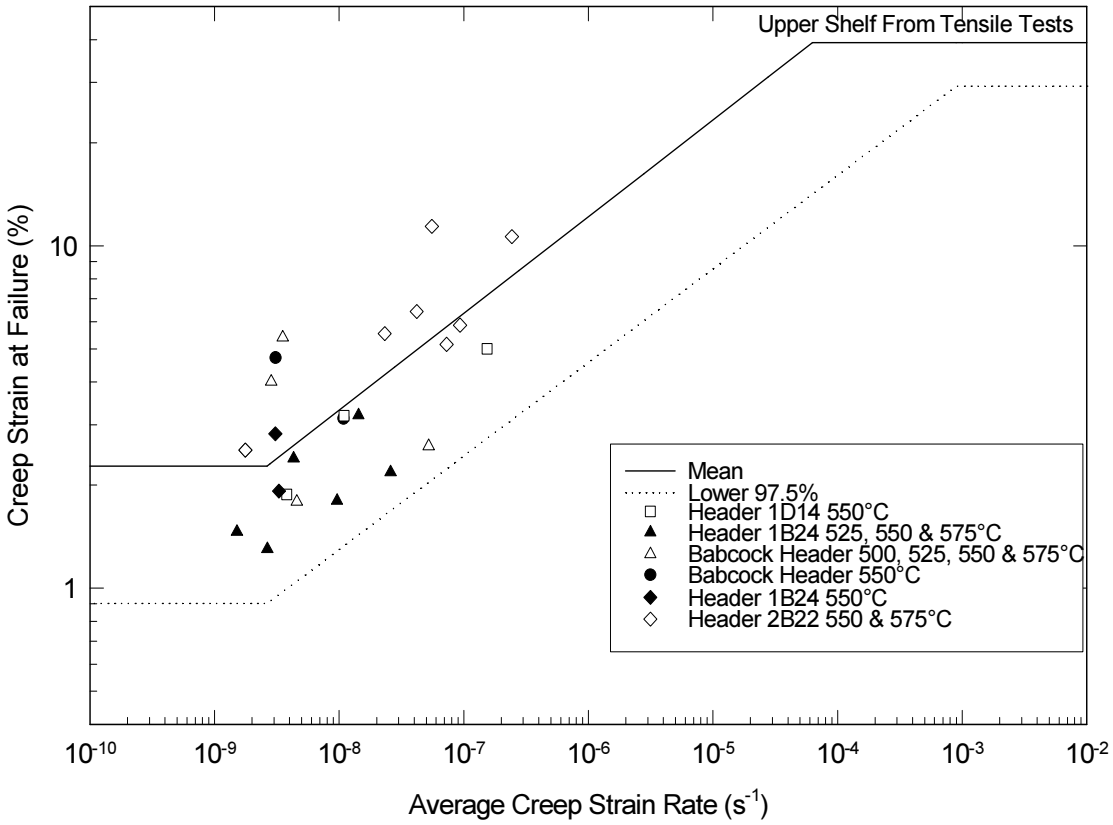
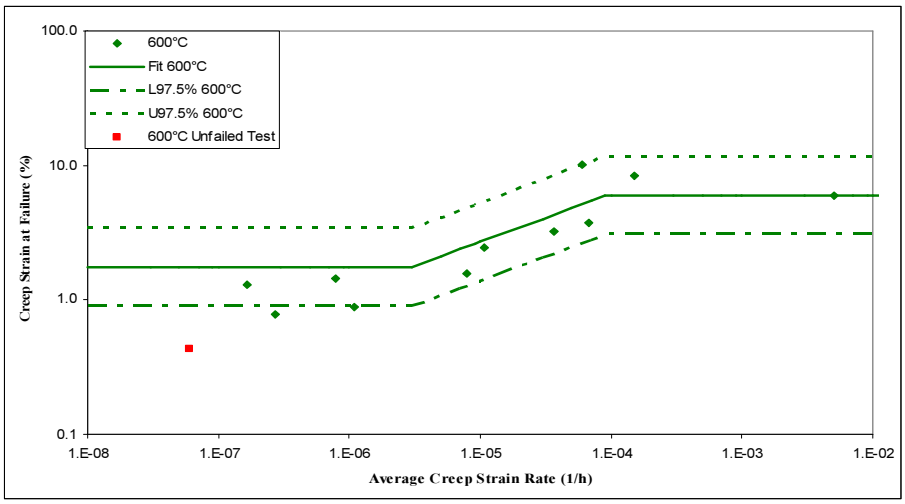
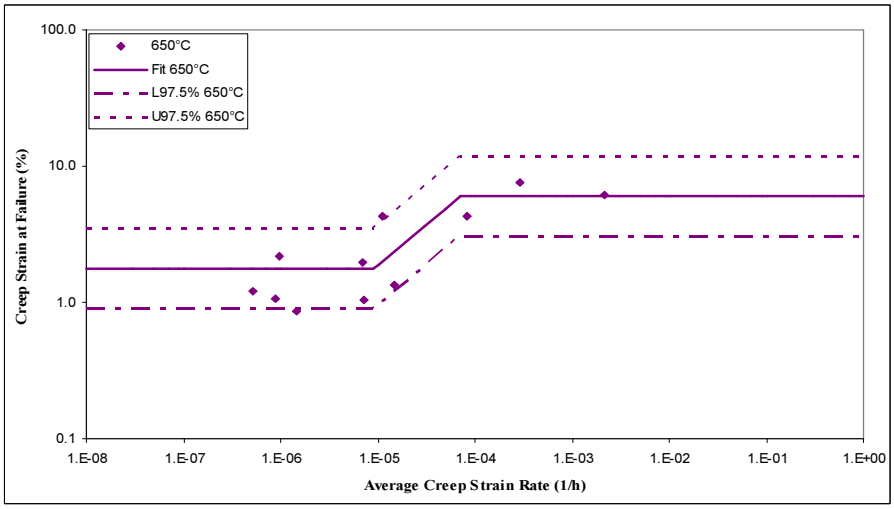


Figure 7b: Last Logged Creep Strain at Failure Data for ESAB OK 69.86 Weld Metal (Esshete 1250) versus the Average Creep Strain Rate cf the Recommended Ductility (i) 600°C and (ii) 650°C (from E/EAN/BBGB/0067/AGR/13).

(i)



(ii)



Qu.: Does creep ductility vary with temperature?

Again this was discussed in Session 23. Materials can have ductility troughs in certain temperature ranges, illustrated for 316/316H below…

Temperature (°C)	Lower 98%CI (%)
500-550	2.6 - 2.8
575	3.3
600	5.6 - 7.4
650	8.4 - 9.4
700	6.6 - 9.0

For this reason 316ss is believed to be susceptible to reheat cracking only below ~550°C.

Qu.: Does creep ductility vary with stress ***state***?

Yes. Very markedly in many cases.

R5V2/3 Appendix A1, Section A1.11.1.2 advises how to account for multiaxial effects on creep ductility, at least for some materials. (Consult the experts, or the References [1-5], for the materials for which this advice applies).

Biaxial Stressing

If one principal stress is virtually zero, and $\sigma_1 > \sigma_2$ are the other principal stresses, the multiaxial Mises ductility, $\bar{\varepsilon}_f$, is,

$$\sigma_2 < 0.5\sigma_1: \qquad \frac{\bar{\varepsilon}_f}{\varepsilon_{f,uni}} = \left(1 - \frac{\sigma_2}{\sigma_1}\right) \tag{5a}$$

$$\sigma_2 > 0.5\sigma_1: \qquad \frac{\bar{\varepsilon}_f}{\varepsilon_{f,uni}} = 0.5 \tag{5b}$$

So biaxial stressing does not decrease the ductility by more than a factor of 2. Note that (5a) implies that the ductility is enhanced by a compressive σ_2.

Triaxial Stressing

For certain austenitic materials, models of cavity nucleation and growth have suggested that multiaxial ductility varies with stress state according to,

$$\frac{\bar{\varepsilon}_f}{\varepsilon_{f,uni}} = \exp\left\{p\left(1 - \frac{\sigma_1}{\bar{\sigma}}\right) + q\left(\frac{1}{2} - \frac{3\sigma_H}{2\bar{\sigma}}\right)\right\} \tag{6}$$

where $\bar{\sigma}, \sigma_1, \sigma_H$ are the Mises, maximum principal, and hydrostatic stresses respectively. Consequently (6) depends upon all three principal stresses.

R5V2/3 Appendix A1, Section A1.11.1.2 includes two sets of values for the parameters p and q. The more onerous values, $p = 2.38$ and $q = 1.04$, were derived from tests on 304 stainless steel. However, these values for the parameters have been widely used for 316H in the reheat cracking temperature range (e.g., Refs.[6-8]). According to R5V2/3 Appendix A1, Section A1.11.1.2 these more onerous values for p and q are deemed to be appropriate in the transition region.

A great deal of work has been done over the last 10-15 years on 316H and Eshhete, for example using notched bar specimens, to obtain data on creep ductility in triaxial states and determine the values of p and q. Consult the experts for the latest position.

Qu.: Mike Spindler's Stress-Dependent Ductility Formulation

Use of the stress-dependent ductility formulation has been sanctioned in the recent revision of R5V2/3 (Issue 3 Rev.002). See also Ref.[9].

The stress-dependent formulation of ductility is inspired by the three distinct mechanism regimes: upper shelf, transition region and lower shelf. The model makes use of the expected, and distinct, functional dependencies on stress, stress state, strain rate and temperature in the three regimes.

(i) Upper Shelf: Cavities grow by power law creep (dislocation controlled);

(ii) Transition Region: Cavities grow by diffusion controlled processes;

(iii) Lower Shelf: cavities grow by diffusion, but diffusion rate is throttled by rate of grain deformation (constrained cavity growth).

See R5V2/3 Issue 3 Rev.002 Appendix A1, section A1.11.1.3 for the functional dependencies of the ductility in each of these regions.

All three regimes are used in fitting the experimental data from uniaxial creep tests. But the upper shelf form is not recommended in applications to plant because the associated necking will generally not occur. Hence, only the transition and lower shelf regions are recommended for plant applications. This gives the multiaxial ductility as,

$$\bar{\varepsilon}_f = MAX\begin{Bmatrix} A_1\dot{\bar{\varepsilon}}^{n_1}\sigma_1^{-m_1}\exp(Q_1/T) \\ MIN\left[\varepsilon_L, A_2\sigma_1^{-m_2}\exp(Q_2/T)\right] \end{Bmatrix}\frac{\bar{\sigma}}{\sigma_1}\exp\left(\frac{1}{2}-\frac{3\sigma_H}{2\bar{\sigma}}\right) \tag{7}$$

The upper term, in A_1, relates to the transition region. Its strain rate dependence assigns larger ductilities at faster strain rates.

The lower term relates to the lower shelf. A stress-dependent ductility fit is thus defined by specifying values for the parameters $A_1, n_1, m_1, Q_1, A_2, m_2, Q_2, \varepsilon_L$, of which only ε_L may be temperature dependent. The temperature dependence of the other two terms is carried by the explicit Arrhenius factors.

Note that the stress dependent formulation gives the multiaxial ductility directly - rather than having to apply a separate reduction factor to account for stress state.

Qu.: How are plastic strains treated in the stress-modified approach?

Plastic strains are retained when the data is fitted to the stress-modified model equations, i.e., the total inelastic strain is used. But the fitting is done using the predicted creep damage, not the strain at failure *per se*. Since the plastic strains accumulate at high strain rate they tend to contribute little damage due to the model assigning high ductilities at fast strain rates (see above).

Qu.: Is the stress dependence of the stress dependent ductility formulation correct?

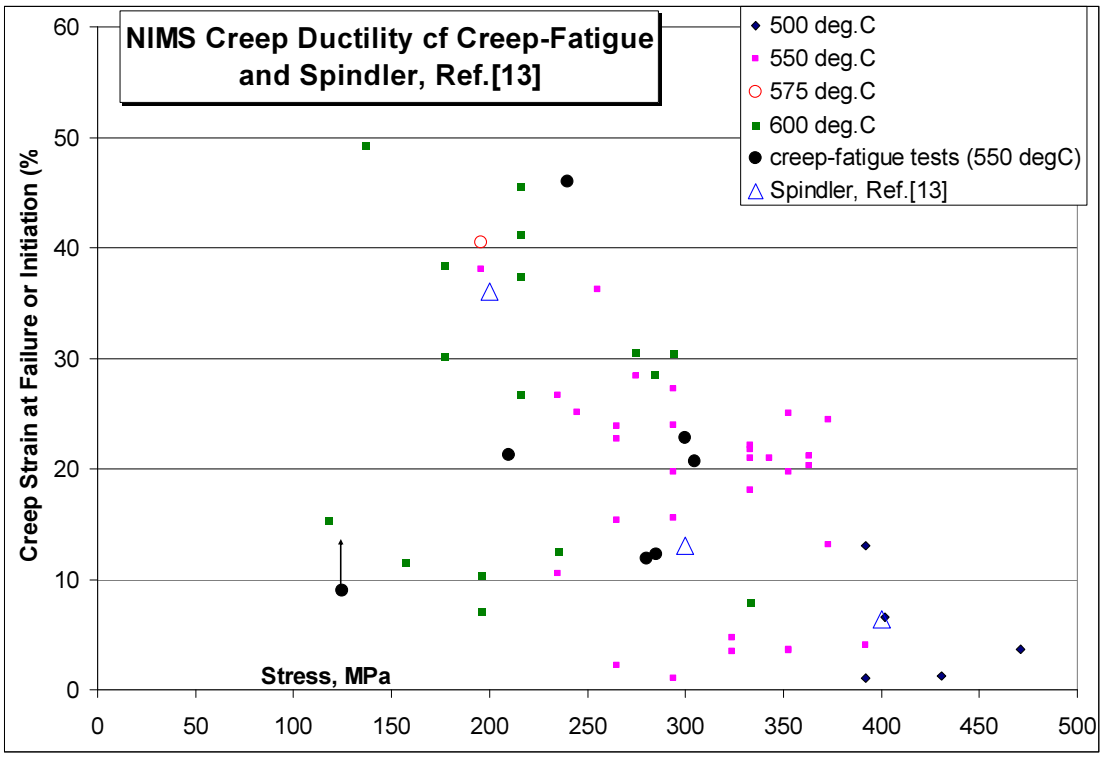
With m_1, m_2 as positive quantities, the stress-dependent ductility, (7), indicates a smaller ductility at higher stresses. This appears (at least superficially) consistent with uniaxial creep test data, e.g., that for 316 parent in Figure 8.

But recall the discussion of session 23 in which I emphasised that higher stresses are strongly correlated with lower temperatures, and vice versa, in the test matrix. So can we really distinguish the effect of stress from the effect of temperature? I have my doubts - but I am a heretic and may soon be burnt at the stake.

Qu.: Materials for which stress-dependent ductility fits are available

A list of internal reports and external publications which provide stress-modified creep ductility fits for a range of materials is given in the References.

Figure 8 316H Creep Ductility and its Dependence on Stress or Temperature (but not clear which!)



Qu.: Is there anything else which affects creep ductility?

How long have you got?

Three things come to mind, all of which can have a major effect on creep ductility,

Prior Plastic Strain

Whilst it is convenient for the analyst to pretend that plasticity and creep are two entirely separate things, in truth they are not. As discussed in session 23, both phenomena are caused by the same crystal defects, such as dislocations, and hence they cannot be independent. Consequently prior plastic straining, e.g., cold work, can have a marked affect upon the subsequent creep behaviour, both the creep strain rate and the creep ductility. Sufficiently large amounts of prior plastic straining can reduce the creep ductility greatly. It is easy to see why: many initially mobile dislocations will be 'used up' (pinned) by the prior plastic straining, and hence become unavailable

for subsequent creep. This phenomenon is a likely contributory factor in reheat cracking. The welding process which creates the residual stresses which drive reheat cracking will also cause large plastic strains – especially in thick sections with many welding passes.

The Stress History

It is sometimes observed that the apparent creep ductility derived from creep-fatigue tests, i.e., under cyclic loading, can be larger than that derived from constant load tests. The reason is that the response of the material depends upon its internal state, and its internal state is a product of its loading history. In other words, the immediately preceding stress states (perhaps a higher stress, or a compressive stress) affects the strain at failure – possibly beneficially, possibly disbeneficially. The stress dependent ductility, equ.(7), can be used to model the effect (in principle, at least) provided non-linear FE analysis has given you the stress.

The Service History

Providing that conditions of stress and temperature do not vary too much, the effects of thermal ageing are implicit in the creep data. However if thermal ageing takes place at modest stress levels, and the structure is subsequently exposed to higher stresses, the creep strain rates may be significantly greater due to the softening caused by the earlier ageing. An effect of thermal ageing on creep ductility is therefore likely. In this example it would be beneficial, i.e., faster creep rates of aged material is likely to be offset by greater ductilities. (This is the behaviour observed, for example, in ex-service tailpipes from HNB).

The Chemical Environment

We have learnt only very recently (since the first version of this lecture was presented) that the AGR reactor CO2 environment will cause surface carburisation (of 9%Cr1%Mo and 300-series austenitic steels, for example). Moreover, this carburised layer will have very markedly reduced creep ductility, as well as reduced plastic ductility and reduced fatigue endurance. Hence surface cracks will be induced far more readily than in nominal material, both by creep and fatigue mechanisms. However the depth of this hardened layer will not be great, perhaps ~0.5mm. So the structural significance of this shallow surface cracking will depend upon the section thickness and/or whether the stresses are sufficient to drive the cracks to greater depths by creep-fatigue. This is an ongoing research area within HiTBASS. One report which has been issued on this topic to-date (Jan'15) is Ref.[10] which is a scoping study on the stress and temperature combinations which will cause such shallow cracks to 'incubate' (i.e., to start to grow). Whether incubation assessments are valid in this context is contentious however.

Qu.: What else does thermal ageing affect?

- The tensile strength,
- The fracture toughness,
- The cyclic stress-strain curve,
- The fatigue endurance,

are all affected by thermal ageing. The toughness of austenitic welds in particular can be severely degraded by ageing at T2 temperatures (sigma phase embrittlement) – but this is not relevant in an R5V2/3 assessment.

The service ageing of the tensile strength and stress-strain properties will usually consist of the combined effects of both thermal ageing and service induced straining.

An instance where the effects of thermal ageing are probably significant in causing creep-fatigue crack initiation is TTIBC in conventional power plant CMV welds. (TTIBC = thermal transient induced bore cracking, a phenomenon which became very widespread in thick section main steam pipework and headers about 10 years ago). Conventional CMV plant in the UK is run at 570-580°C. The resulting softening leads to larger strain ranges on thermal cycling. This together with the deleterious effect on fatigue endurance is the cause of the cracking, a form of thermal fatigue (recall also that conventional plant is subject to far larger numbers of cycles due to two-shifting).

R5V2/3 Appendix A1, Section A1.5.3.1 suggests that an FSRF of ~1.5 is required to account for thermal ageing in 2.25Cr1Mo after very modest amounts of ageing at temperatures appropriate for our steam plant. I don't know the provenance of this advice, nor have any experience of its implications. Does anyone use this FSRF?

For more on thermal ageing effects see R66 and consult the experts.

Qu.: What is the significance of "continuous cycling" fatigue endurance data?

A fatigue endurance test can be carried out with a (roughly) sinusoidal variation of load with time. This would mean that the peak load occurs only instantaneously. This is a "continuous cycling" test.

Alternatively the peak load may be held for some period (the "dwell" time).

At temperatures within the creep regime, it makes a great deal of difference whether a fatigue endurance test includes dwells or is continuous cycling. The reason is simply that a dwell at high stress at creep temperatures will induce creep damage in addition to the fatigue damage. So a fatigue test with dwells is just another name for a creep-fatigue test.

Since R5V2/3 assessments account separately for the creep and fatigue components of damage, it would be double accounting to use fatigue tests with dwells as the source of "fatigue" endurance data. Consequently, continuous cycling fatigue endurance data are required in an R5V2/3 assessment. *(It is worth noting that in R5V4/5 assessments of creep-fatigue crack growth the corresponding issue is rather more involved – see later sessions).*

Qu.: How is creep deformation formulated?

Qu.: What is meant by the hardening law in creep?

For these issues see Session 24.

Qu.: How is creep relaxation evaluated?

This will be the subject of a session of its own later, Session 34.

Qu.: What is the "elastic follow-up factor, Z"?

The R-procedures, like design codes, are based on the simplifying notion that loads can be classified as primary or secondary. But actually loads are sometimes of an intermediate character. An example is an applied displacement or rotation. If such displacement controlled loads are applied over a small gauge length, they may approximate to a secondary load. On the other hand, if applied over a very long gauge length they may approximate to a primary load. In general, therefore, they are of intermediate type. R6 contains specific advice on the treatment of displacement controlled loads for this reason.

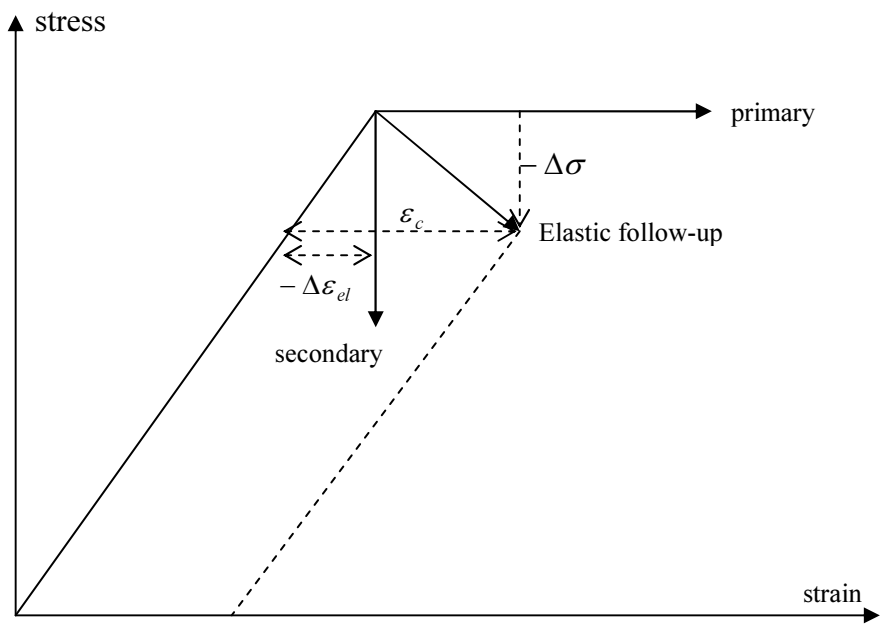
In creep the distinction between primary and secondary loads takes the following form: under primary loading the creep strain accumulates at constant stress, whereas under a pure secondary load the creep strain accumulates at a constant *total* strain, whilst the stress relaxes. In the latter case, creep converts (part of) the initial elastic strain into creep strain, i.e., $\varepsilon_c = -\Delta\varepsilon_{el} = -\Delta\sigma / E$.

Elastic follow-up is a way of talking about how loads of intermediate character behave. The stress does relax (unlike primary loading), but it does not relax as much as it would under pure secondary loading. On the other hand, a greater amount of creep strain accumulates over a given time than would be the case if the load were purely secondary. The elastic follow-up factor, Z, is the factor by which the creep strain exceeds the reduction in elastic strain:-

$$\varepsilon_c = -Z\Delta\varepsilon_{el} = -Z\Delta\sigma / E \tag{8}$$

There is an important variant, or limitation, to this definition which will be discussed in detail in Session 34. However, for now consider Figure 9…

Figure 9: Definition of Z



Qu.: Is the creep strain Z times what it would be if Z=1?

No.

It is tempting to conclude this from equ.(8) but it should be noted that the stress drop, $\Delta\sigma$, is less if $Z > 1$, so the creep strain is less than Z times what it would be in pure secondary loading (i.e., if $Z = 1$).

Qu.: What is the exact definition of creep damage in R5V2/3?

The notes above implied that $D_c \approx \dfrac{\varepsilon_c}{\varepsilon_f}$ is only approximate. This is because the full expression must involve an integral over varying conditions, because the creep ductility varies with strain rate and with stress state – both of which will change over time. So the correct expression is,

$$D_c = \int_0^t \frac{\dot{\bar{\varepsilon}}_c dt}{\bar{\varepsilon}_f\left(\dot{\bar{\varepsilon}}_c, \sigma_{ij}\right)} = \int_0^t \frac{\dot{\bar{\varepsilon}}_c dt}{S\left(\sigma_{ij}\right)\bar{\varepsilon}_{f,uni}\left(\dot{\bar{\varepsilon}}_c\right)} \tag{9}$$

In (9) the numerator involves the Mises equivalent creep strain rate. The denominator is the multiaxial creep ductility evaluated for the instantaneous strain rate and stress state. In the second form of expression this multiaxial ductility has been expressed as the product of a strain rate dependent uniaxial ductility and a stress state dependent Spindler fraction.

Qu.: Can R5V2/3 be used to assess reheat crack initiation?

Yes.

But you must remember not to confine yourself to surface points. Stresses can only be biaxial at worst at a free surface. So triaxiality necessarily peaks sub-surface. Not surprisingly, then, FE models of reheat cracking predict initiation sub-surface (consistent with plant experience in many cases - though surface cracking is possible at a sharp weld toe feature).

And, of course, you will need to know the residual stresses in sufficient detail. This means knowing *all* components of stress. This is because, as we saw above, even the smallest of the principal stresses is crucial to the Spindler fraction and hence to crack initiation. So predictions of reheat cracking require a more detailed knowledge of the full stress tensor than one might often get away with in assessments. For this reason, FE modelling of the welding process is generally required. Once such a model exists, it is little further effort to implement the creep damage model, i.e., (9), in FE also.

Qu.: Are "by hand" assessments of reheat crack initiation feasible?

It is possible to assess simply whether reheat cracking is, or is not, likely to be a threat – but only if you can anticipate the degree of residual stress triaxiality. The following applies to 316H parent welded in the usual manner with a 316 consumable and not subject to PWHT. The reheat crack initiation criterion can be written roughly as $D_c \approx \dfrac{\varepsilon_c}{S\varepsilon_{f,uni}} = \dfrac{Z|\Delta\varepsilon_{el}|}{S\varepsilon_{f,uni}}$. For 316H the lower bound, lower shelf creep ductility is usually taken in reheat cracking assessments to be $\varepsilon_{f,uni} \approx 1\%$. The stress drop is likely to be (again roughly) of the order of the proof strength, and so $|\Delta\varepsilon_{el}| = |\Delta\sigma| / E \approx 0.1\%$, because the elastic strain at the 0.2% proof stress is typically

about 0.1% for 316H. This gives $D_c \approx \frac{Z \times 0.1\%}{S \times 1\%} = \frac{Z}{10S}$. Consequently damage reaches unity if $\frac{Z}{S} = 10$. So our crude assessment of the risk of reheat cracking in 316H is,

Reheat cracking of a weldment in 316H is likely if, at the most onerous point, $\frac{Z}{S}$ exceeds 10. Otherwise reheat cracking will probably not initiate.

Data from detailed FE analyses, Figure 10, confirms that this is a good estimator of the risk of reheat cracking. However this 'rule-of-thumb' does not carry the same authority as a full analysis of the weld in question. Nor is it necessarily indicative for other materials.

Figure 10: Simple Reheat Cracking Risk Assessment (for 316H Welds)

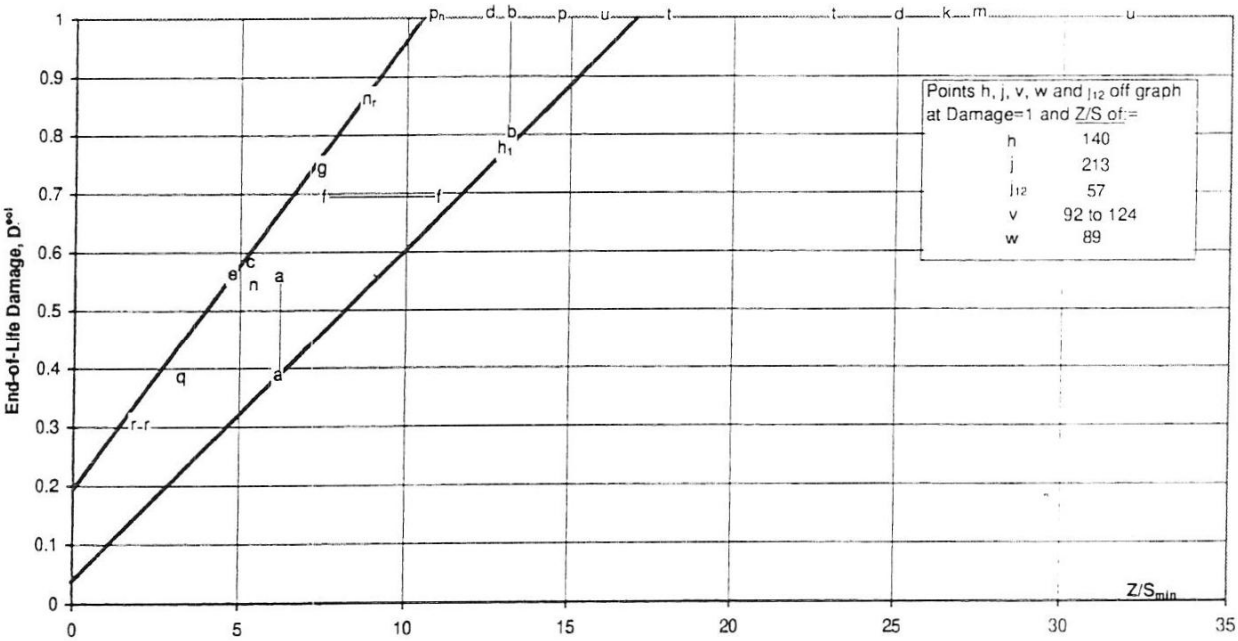


Figure 5: Damage versus Reheat Cracking Parameter, Z/S_{min}

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