T73S06 Session 28: Transition Joints (R5V6)

Addresses T73S06 items 6.1 to 6.8, plus items in T73S04

Last Update: 8/12/14

Types of transition joints and where are they?; What is the unique TJ cracking mechanism?; R5V6: is it failure or initiation? R5V6 A, B & C terms; Miner's Law; FSRFs; $\Delta\alpha\Delta T$ *effect and distinction between contributions to B & C; Methodology for strain range and endurance data for term B; creep data to be used for term C; Is the* $\Delta\alpha\Delta T$ *damage real? Salutary lesson re* $\Delta\alpha$ *(HPB/HNB UTJs).*

Do not use these notes as a source in safety related assessments – UNVERIFIED!

Qu.: What is a "Transition Joint"?

Strictly the scope of R5V6 is "dissimilar metal welds" (DMWs). These are any welds where the two parent materials being joined are sufficiently different in terms of either physical, mechanical or creep properties.

However, in practice, the application of R5V6 is always restricted to transition joints, which are defined as weldments connecting ferritic and austenitic parent materials.

Qu.: How many types of TJ do we have in our plant?

Lots.

Formally the complete inventory (which is obligatory under TGN046, Ref.1) is the Weld Inventory Database (WID). However you may find Tables 1-4 in Ref.2 more convenient. This is rather an old reference, so it is possible that some of the TJs have been replaced by now. But it gives a good overview of the types of TJ which have been used. Below is an incomplete list to give some rough idea. The list is in order of creep life, starting with "excellent" and ending with "crap":-

Graded Joints (Jessop-Saville): These TJs involve a gradual change in material composition from 2.25%Cr1Mo to 316ss, thus avoiding the problems which other TJs suffer (carbon diffusion and locally high thermal stresses due to a sharp discontinuity in α). Used at DNB, HY2, TOR and HNB.

Stress-Relieved Inconel Filled: All reviews have concluded (based on plant experience) that these nickel-based filler welds have better creep life than austenitic filled welds (see Refs.2,3) by a factor of at least 2 and typically more like 3 or 4. These have been used as replacements for earlier TJ designs at HYA/HRA/HPB.

English Electric Mark III: These connect 2.25%Cr1%Mo to 316ss via a 316ss weld filler. The ferritic parent is first buttered with a niobium stabilised layer of 2.25%Cr1%Mo weld material, followed by buttering over this with 316ss weld material. Then the main 316ss weld is made, and the complete joint 'conditioned' at 600°C. This hardly constitutes a heat treatment as such. The purpose of the niobium stabilised layer is to inhibit carbon diffusion and hence attempt to avoid the TJ cracking mechanism described below. These are used at HPB and DNB. There was some scepticism in the early 90s as to whether the HPB HRH EEMkIIIs would last the course, and some were replaced (with inconel filled TJs). However, I suspect that operating experience of those not replaced has been OK so far (2014) – (see Ref.1).

HPB/HNB UTJs: These are the upper transition joints in the boilers, connecting the top of the 9%Cr1%Mo primary superheater platens to the bottom of the 316H secondary superheater platens. The joint consists of two welds. The upper weld connects 316H to a Sanicro 71 (=Inconel 600) transition piece using a 316 filler. The lower weld connects the Sanicro to 9%Cr1%Mo using a 9%Cr1%Mo filler. The use of a ferritic filler in the lower weld makes these joints unique. (I have been unable to identify the use of a ferritic filler in any other TJs, including surveys of non-BE plant). On the face of it this design seems like a good idea. The coefficient of thermal expansion of Sanicro is intermediate between that of 316 and ferritic steel. Accordingly, the intention was to reduce the magnitude of $\Delta\alpha$ across any interface, and hence minimise the thermal stresses. Unfortunately measurements of the α of material taken from the lower weld (nominally 9%Cr1%Mo filler) has shown it to be unexpectedly high – even higher than 316 parent, if anything. The reason would appear to be that so much Ni has leached out of the Sanicro into the weld pool that the filler has effectively been turned into an austenitic material. So there is a severe mismatch thermal stress contribution to the R5V6 assessment of these welds. (There was also a solidification cracking issue with these welds – but that's a separate issue). On the other hand, as of 2014 (after 38 years operation) no UTJ leaks have definitely occurred on any of the four HPB/HNB reactors, although there have been 3 leaks that could possibly be from UTJs. So there is no plant difficulty with these TJs at present.

<u>Non-Stress-Relieved Inconel Filled</u>: These joints were made by initially welding two ferritic 2.25%Cr1Mo pipes together with the Inconel filler, and then stress relieving this weld. The joint was then cut down the weld centre line and a new weld prep made in the Inconel, thus leaving a layer of Inconel weld material on the ferritic parent. The TJ was then made to the 316ss using the same Inconel filler, but with no subsequent heat treatment. There was a sorry history with this design of TJ, which was used initially to connect the 316H superheater outlet header nozzles at HYA/HRA to the ferritic main steam pipework. A leak occurred after only 8000 hours service, and inspection revealed extensive cracking of the other welds. All were replaced. The cause of this very early cracking was never fully determined. Residual stresses would appear to be one of the main causes, but it might be that there were other contributory factors. Because of this experience, some caution is needed as regards the implications of residual stresses for TJs. This caution should (in my opinion) extend to the applicability of R5V6 (see below).

There are lots of other TJs not covered in the above list, e.g., UTJ designs at other than HPB/HNB; LTJs at all Stations; various TJs on the HYA/HRA boiler spines; various TJs on penetrations (including CMn-316 joints) – see Ref.2 Tables 1-4.

Qu.: What is the scope of R5V6 in terms of TJ Type?

The scope of R5V6 is far more restricted than other parts of R5. Not only is it limited to DMWs, but in practice it covers only circumferential butt welds in pipes made with a very limited range of materials. The limitation to circumferential butt welds is not as restrictive as it might have been because, in our plant at least, TJs are always of this type – although in a wide range of diameters and thicknesses.

But unfortunately we do not have the necessary cross-weld materials data for many of our TJs. What we do have is,

parent	weld	parent	examples
2.25%Cr1%Mo	316	300-series	EE Mk.III (HPB/DNB)
2.25%Cr1%Mo (PWHT'd version)	Nickel alloy	300-series	HYA/HRA/HPB header to pipework replacements
Annealed 9%Cr1%Mo	Inconel	316	HYA/HRA UTJs
Normalised & tempered 9%Cr1%Mo	Inconel	316	HRA boiler spines
Normalised & tempered 9%Cr1%Mo	Inconel	Alloy 600	HY2/TOR UTJs
9%Cr1%Mo	9%Cr1%Mo	Sanicro 71	HPB/HNB UTJs
Jessop-Saville			Main steam pipework DNB, HY2, TOR, HNB

As usual, check the User Queries lists (both R5 and R66) for updates on TJ data. However, I suspect that Section 6 of R66 Rev.009 (2011) is probably the current advice.

Use only R66 and the User Queries, plus expert guidance, for TJ rupture.

Do not use the data in R5V6 Issue 3 (2003) Appendix A1.

In particular the 9Cr1Mo-Inconel-316 data in R5 Issue 3 (2003) is wrong.

Qu.: What makes TJs different?

There are two things which make TJs very different from similar metal welds.

Mismatch of Thermal Expansion Coefficients

Ferritic and austenitic steels have very different coefficients of thermal expansion. Consequently there is a mismatch, $\Delta\alpha$, over the joint. This gives rise to a large local thermal stress even when the temperature is uniform. This effect is included in the R5V6 assessment procedure.

Specific TJ Cracking Mechanism

TJs are potentially subject to a failure mechanism which does not affect similar metal welds. This involves carbon diffusion out of the ferritic material near the fusion boundary and the formation of carbide precipitates close to the fusion line (described further below). Protection against this failure mechanism should, in principle, be provided by the R5V6 procedure by virtue of it being based on cross-weld rupture data. Should this cracking mechanism be relevant under the assessed conditions, allowance for this would then be implicit in the data. But this is dependent upon the test data being of sufficient duration, and not accelerated too much by uncharacteristically high stresses (see below).

Qu.: What causes this TJ-specific cracking mechanism?

A mixture of free chromium and free carbon in solid solution is not thermodynamically stable. A reduction of free energy can be achieved if the Cr and the C react to form carbides (typically $Cr_{16}Fe_7C_6$). But carbide precipitation tends not to occur within the crystal matrix because there is no room. Instead it occurs on grain boundaries where the disruption of the crystal structure creates some space. At high temperature, carbon diffuses readily through the Fe matrix. Consequently the Cr from the austenitic or Inconel weld filler reacts with the carbon from the ferritic steel at, or very near, the fusion boundary. Generally these carbides precipitate just one or two microns away from the fusion boundary (typical for nickel based fillers), this being a small fraction of the grain size (typically 40-200 microns). The decrease in concentration of free Cr and free C in this region, due to being used up by the reaction, causes more Cr and C to diffuse into the region from the rest of the affected grains. The resulting line of carbides creates a low ductility path for cracking parallel to the fusion line.

In some cases the crack path might be a little further from the fusion boundary, but still within one or two grains. This is especially the case with austenitic fillers. In these cases the carbide precipitation has occurred on the "prior austenite-grain boundaries" (PAGB)*. The distinction does not appear particularly important to the analyst, as far as I can see.

*This is a potentially confusing bit of terminology. This is what I think it means. It refers to the fact that at the high temperatures which can occur in the HAZ during welding, or during PWHT, austenite will tend to form. But this austenite phase is unstable and reverts back to ferritic phases as the metal cools. However, the position of the austenite grain boundaries prior to this re-conversion can still be seen in the ferritic metal when cold – and it can be these "prior austenite grain boundaries" which attract the carbide precipitates.

Qu.: What implications does this cracking mechanism have for R5V6?

It means that the cross-weld data used in an R5V6 assessment must be obtained from tests in which this mechanism was permitted to occur, if it was going to. In particular, this means that the test duration must be long enough for carbon diffusion and carbide precipitation to take place. In practice this means not using too high a stress, and ideally using representative temperatures. Tests conducted at high stress tend to exhibit good ductility and a strength/creep life consistent with the parent ferritic material. However, inferior strength/creep life, and substantially reduced ductility, compared with parent occurs at lower stresses and longer testing times. It is the latter which are most often relevant to plant.

Qu.: Does R5V6 assess crack initiation, crack growth or gross failure?

Crack growth is *not* covered in R5V6. For this see R5V4/5.

R5V6 assumes the joint is defect-free.

The distinction between crack initiation and gross failure is not made terribly clear in the procedure. But ***R5V6 really assesses gross failure***, because:-

- R5V6 does not make any explicit mention of crack initiation;
- Unlike R5V2/3 Appendix 10, there is no correction of fatigue damage for the size of any initiated defect;
- Creep damage term A is calculated with respect to gross rupture data;
- The creep-fatigue damage mechanism is not addressed, i.e., elastic cycling is implicitly assumed. R5V6 is a little naughty in not making this restriction explicit.
- Apart from the $\Delta\alpha$ effect, secondary stresses are not included in the creep damage part of R5V6 (but they are included in term B, the fatigue damage part).

Qu.: Did you say that R5V6 was for elastic cycling only?

Yes.

It is not stated in the procedure, but R5V6 assumes that the load cycling is elastic. The procedure is rather schizophrenic because it talks of plasticity corrections to the strain range when calculating the fatigue damage – but this is inconsistent and can be ignored.

Qu.: What should be done if plastic cycling is important?

Qu.: What should be done to assess crack initiation in a TJ?

R5 does not provide a procedure in these cases.

However, you can be creative and knock something together from a combination of R5V2/3 and R5V6 – it's been done before. Consult the experts!

Qu.: How is an R5V6 assessment structured?

The assessment calculates three contributions to damage, *A, B* and *C*, so that the total life fraction is $A+B+C$, where,

A = Creep damage due to primary stresses;

B = Fatigue damage;

C = Creep damage due to α mismatch.

Note that R5V6 treats system loads as primary. This is consistent with our normal policy for pipework system loads, but for TJs the procedure actually requires it.

Qu.: How is creep rupture (term A) calculated?

Term A is the total primary-stress creep damage based on Robinson's Rule,

$$A = \sum_i \frac{t_i}{t_{rup}\left(\sigma_{rup,i}, T_i\right)} \tag{3}$$

where t_i is the total time spend at temperature T_i and conditions giving a representative rupture stress of $\sigma_{rup,i}$, and t_{rup} is the rupture time under these conditions based on cross-weld tests.

- The cross weld test data is given in R66 Section 6 (do not use R5V6 App.A1).
- R5V6 Appendix A2 defines the procedure for calculating the representative rupture stress, σ_{rup}. This procedure is unique to TJs. σ_{rup} is *not* equal to the reference stress or the rupture reference stress used in R5V2/3 and R5V7. (A reference stress method is used to find an effective axial stress, but this is not the stress which is plugged into the rupture equation. It is just an intermediate step). There is little point in reproducing the equations from Appendix A2 here.
- The R5V6 Appendix A2 method covers pressure loading, end loading and system loading. System loading is therefore treated as primary.
- The R5V6 Appendix A2 method explicitly includes allowance for the effect of stress multiaxiallity on creep rupture. (This contrasts with, say, R5V2/3).

Qu.: Is it necessary to correct σ_{rup} in R5V6 for a local stress concentration?

No.

The question relates to the distinction in R5V2/3 and R5V7 between the reference stress, σ_{ref}, and the rupture reference stress, σ_{ref}^R, the latter being factored by an expression depending upon the SCF, χ. But there is no such requirement in R5V6 because the rupture data and the representative rupture stress have been obtained empirically from tests on specimens with the relevant weld types.

Qu.: How is fatigue damage (term B) calculated?

Miner's Law is used to sum the fatigue damage due to different cycle types,

$$B = \sum_j \frac{n_j}{N_f\left(\Delta\bar{\varepsilon}_j\right)} \tag{4}$$

where n_j is the number of cycles of type j, each of which produces a Mises equivalent strain range of $\Delta\bar{\varepsilon}_j$ with an associated endurance of $N_f\left(\Delta\bar{\varepsilon}_j\right)$.

Both primary and secondary cyclic stresses contribute to term B, e.g., due to pressure, system loading and thermal effects - including α mismatch.

Qu.: How is the equivalent strain range $\Delta\bar{\varepsilon}_j$ calculated?

The strain range is the sum of two parts: that which would be calculated for a homogeneous body following R5V2/3 Appendix A7, $\Delta\bar{\varepsilon}_{V2/3}$, and that due to the mismatch in the coefficient of expansion, $\Delta\bar{\varepsilon}_t$. So,

$$\Delta\bar{\varepsilon}_j = \Delta\bar{\varepsilon}_{V2/3} + \Delta\bar{\varepsilon}_t \tag{5a}$$

where,
$$\Delta\bar{\varepsilon}_{V2/3} = \Delta\bar{\varepsilon}_{el} + \Delta\bar{\varepsilon}_{\mathrm{pl}} + \Delta\bar{\varepsilon}_{vol} \tag{5b}$$

and,
$$\Delta\bar{\varepsilon}_t = 1.5\Delta\alpha\Delta T_j \tag{5c}$$

The three terms in (5b) are explained in R5V2/3. In particular, the plastic strain range, $\Delta\bar{\varepsilon}_{\mathrm{pl}}$, requires construction of the hysteresis cycle following the procedure of R5V2/3 Appendix A7. This is complex and will be dealt with in future sessions on T73S04.

> However, since R5V6 is actually restricted to elastic cycling, (5b) becomes simply $\Delta\bar{\varepsilon}_{V2/3} = \Delta\bar{\varepsilon}_{el}$. This is a glitch in the procedure. Fortunately it means that those seeking SQEP accreditation in T73S06 do not need a large part of T73S04 first.

Qu.: What does the α-mismatch term, (5c), mean?

The novel contribution to the strain range is the thermal strain due to the α mismatch, $\Delta\bar{\varepsilon}_t$, given by (5c).

In this term, $\Delta\alpha$ is the difference in the coefficient of expansion over the weld fusion boundary on the side of the weld being assessed. Generally the creep rupture term will be most significant on the ferritic side, so that it will be the ferritic side which is assessed. In this case $\Delta\alpha$ is the difference between the weld material α and the ferritic parent α.

The temperature difference, ΔT_j, relates to the cycle type j in question. It is the maximum difference in the weldment temperature over the cycle. Note that the cycle might be sufficiently slow that the temperature is virtually uniform at all times – and hence would not produce any thermal stress in a homogeneous body. However the α mismatch still causes a thermal strain range of $\Delta\bar{\varepsilon}_t = 1.5\Delta\alpha\Delta T_j$.

Note that ΔT_j may not be maximum at the same time that $\Delta\bar{\varepsilon}_{V2/3}$ is maximum. In this case it would be sensible (in my opinion) to use the maximum value of the sum $\Delta\bar{\varepsilon}_j = \Delta\bar{\varepsilon}_{V2/3} + \Delta\bar{\varepsilon}_t$ at any time during the cycle, rather than adding the maxima of the two terms. However, R5 does not make this explicit.

Amongst other possible cycles, term B will include start-up/shut-down cycles (which will inevitably have a large value for ΔT_j).

Qu.: Why $\Delta\bar{\varepsilon}_t = 1.5\Delta\alpha\Delta T_j$?

Clearly the elastic stresses due to the α-mismatch will be proportional to $\Delta\alpha\Delta T_j$, but why the factor of 1.5? This was based on finite element work reported in 1993 by Royden Hales & John Phillips, TIGM/MEM/0092/93. I wouldn't die in a ditch defending it.

Qu.: Do thermal transients contribute to $\Delta\bar{\varepsilon}_j$?

The procedure states that thermal shock loadings are not covered.

However thermal stresses and primary stresses contribute in the same manner to the fatigue damage, term B – see Section 5.3 and Appendix A3, Section A3.1. The thermal loads must be included in the calculation of $\Delta\bar{\varepsilon}_{V2/3}$, and hence of $\Delta\bar{\varepsilon}_j$.

The fact that austenitic steels have a significantly smaller thermal conductivity and a larger coefficient of thermal expansion compared with ferritic steels is relevant since this will enhance any thermal transient stress, as would any thickness mismatch.

Qu.: Are thermal transient cycles assessed in addition to the $\Delta\bar{\varepsilon}_t = 1.5\Delta\alpha\Delta T_j$ cycles?

To be safe, yes.

However, a fatigue assessment requires only that the maximum and minimum total stress during the cycle be identified, in order to find the correct stress range. Whilst transient thermal stresses may arise, this does not mean that they necessarily make any contribution to the stress range or the fatigue assessment. The conditions leading to the maximum α mismatch thermal strain range, $\Delta\bar{\varepsilon}_t = 1.5\Delta\alpha\Delta T_j$, are quite likely to arise only after the thermal transient stresses have died away (though this would need to be confirmed in any given case). And because $\Delta\bar{\varepsilon}_t = 1.5\Delta\alpha\Delta T_j$ is generally a large strain, it will often be found that $\Delta\bar{\varepsilon}_j = \Delta\bar{\varepsilon}_{V2/3} + \Delta\bar{\varepsilon}_t$ continues to define the correct maximum strain range despite the transient stresses.

Qu.: What is the Fatigue Strength Reduction Factor (FSRF)?

Weldments are more susceptible to fatigue damage than parent material. This is due to a combination of factors.

Firstly the geometry of a weld can exacerbate fatigue, especially if the weld cap is still present. The stress concentration at a weld toe feature will generally have a very marked effect. This is where a crack will initiate.

Secondly, the weld and HAZ material can differ from the parent, so that there is also a material effect.

Both effects are included in the FSRF, which is obtained by testing welded specimens or features. The FSRF is defined as the ratio of the plain parent strain range to the welded feature strain range giving the same endurance.

Qu.: Is the FSRF used to factor the stress range or the strain range?

The FSRF is a factor on strain range, *not* on stress or cycles.

Qu.: Is an FSRF necessary in evaluating term B?

If the fatigue endurance, $N_f\left(\Delta\bar{\varepsilon}_j\right)$, is obtained as the lower bound of data from cross-weld tests on the relevant type of TJ, then no FSRF is required. It is already implicit in the test data.

Frequently such cross-weld fatigue data will not be available. In this case, lower bound parent fatigue endurance should be used, together with an FSRF of 2.

Qu.: What is Term C?

Term C measures the creep damage due to the relaxation of secondary stresses due to the α mismatch. It is envisaged that the operating temperature may vary, so that there are effectively n_j cycles for which the operating temperature changes by ΔT_j between periods of steady temperature. It is assumed that the dwell period after each such temperature change is sufficiently long, and the temperature is sufficiently high, that the associated thermal stress is fully relaxed. The strain increment is therefore taken to be $\Delta\alpha\Delta T_j$. Term C is thus,

$$C = \sum_j \frac{n_j \Delta\alpha\Delta T_j}{\varepsilon_f} \qquad (6)$$

where ε_f is the creep ductility.

Note that the temperature cycles contributing to term C need not be related to the fatigue loading cycles. Consequently the temperature ranges, ΔT_j, used in terms B and C will be different in general. Typically the ΔT_j for term B (fatigue) will be large, whereas those for term C will be small.

In particular, the major start-up/shut-down cycles (for which ΔT might be greater than 500°C, say) will invariably contribute to fatigue (term B), but should ***not*** be included in term C, for which the contributing ΔT will typically be a few degrees or a few tens of degrees.

Qu.: What ductility, ε_f, is used in term C, Equ.(6)?

Generally the lower shelf ***uniaxial*** creep ductility of the ***ferritic parent***.

Thankfully R5V6 does not require either the ductility of the weld material, or HAZ, nor does it call for multiaxial effects to be taken into account. Instead the *lower shelf* creep ductility of the parent material is recommended. Assuming that the ferritic side is being assessed, then ε_f will be the ductility of the ferritic parent. Note that the lower shelf ductility is what prevails at very slow strain rates. In general a lower bound to available data should be used for ε_f.

Where specific advice for the ferritic parent ductility is not available, R5V6 Appendix A4 helpfully sanctions the use of $\varepsilon_f = 5\%$.

Qu.: Why is there no factor 1.5 in term C as there is in term B?

I don’t really know.

But given the vagueness surrounding the creep ductility used, a factor of 1.5 would probably suggest a spurious level of precision in any case.

My own rationalisation is that, being a contribution to creep damage, something more like a reference stress is required, rather than the peak stress which is relevant in fatigue. This would motivate a 2/3 reduction factor for a bending dominated stress distribution. This argument may well be rubbish but it's the best I can do.

Qu.: Why are start-up/shut-down cycles not included in Term C?

The implicit assumption of R5V6 is that the loading cycles give rise to elastic cycling, not hysteresis loops. Consequently the stress-strain graph is assumed to be like R5V2/3 Figure A3.5(b). Apart from a small amount of relaxation damage in the first few cycles, the creep damage is then based on creep at a constant (primary) stress (i.e., the rupture term A). To be thorough, there should be an allowance for the initial transient. Using $\Delta\alpha\Delta T / \varepsilon_f$ for this might seem reasonable, where ΔT is the start-up/shut-down temperature difference. However, if the joint were stress relieved, the stress $E\Delta\alpha\Delta T$ would occur when cold. The hot stress would be relatively small and hence not cause creep strain. In any case, there would be just one cycle producing damage $\Delta\alpha\Delta T / \varepsilon_f$. The whole history of reactor start-up/shut-down cycles would not count, since these cycles are assumed to be elastic and not associated with additional creep damage.

Qu.: Could other secondary stresses be incorporated in the same manner?

Suppose there was (say) a through-wall temperature gradient giving rise to a thermal stress. This must be included in the fatigue damage (term B) but is not included in the creep damage according to the R5V6 procedure. By the same argument as above, elastic cycling means that only one relaxation cycle contributes, so it is unlikely to be a significant contribution to the damage. However it can easily be included in term C as an additional contribution $\sigma^{th} / E\varepsilon_f$ if you are concerned about it.

Qu.: Should term C be evaluated for the austenitic side of a TJ also?

The assessment of a TJ should cover both sides. But usually it is the ferritic side which is limiting, because this will have the larger term A, which is likely to dominate term B.

On the other hand, term C could be larger on the austenitic side because of the poorer creep ductility. And in this case it becomes more questionable to ignore stress triaxiality effects on the ductility.

But, to be pragmatic, the TJ cracking mechanism which we are most concerned to guard against, occurs on the ferritic side. So generally it is reasonable to concentrate on the ferritic side.

Qu.: How does R5V6 address PWHT / residual stresses?

It doesn't.

There is no methodology in R5V6 for assessing residual stresses. The joints are assumed to have been "stress relieved" or "temperature conditioned", or whatever other process is required by the relevant design code. TJs without any form of heat treatment would not be covered by R5V6. The procedure is remiss in not making this explicit.

This means that R5V6 would not be applicable to the original design of s/h outlet header to main steam pipework TJs used at HYA/HRA. As noted above, these joints cracked within the first year or two of operation.

Qu.: How does an R5V6 assessment differ from an R5V2/3 assessment?

- It assesses failure, not crack initiation;
- The stress used in creep rupture is different;

- Cross weld data is an essential input to the creep rupture assessment;
- There is no size correction in the fatigue assessment;
- The FSRF advice is different;
- R5V6 is restricted to elastic cycling. It does not cover TJs undergoing stress-strain hysteresis cycles, i.e., the creep-fatigue mechanism is not addressed;
- Consequently, the strain range used in the fatigue assessment does not need to follow the R5V2/3 weldments methodology of Appendix A4 (phew!);
- The strain range used in the fatigue assessment is Mises, not Tresca or Rankine, and there is no mention of multiaxial effects in fatigue;
- There is a $\Delta\alpha$ mismatch contribution to both fatigue and creep damage;
- Term C uses the uniaxial creep ductility, not the multiaxial ductility;
- Secondary stresses other than the α-mismatch stresses do not contribute to creep damage (but note that pipework system loads are included as primary stresses). However they can easily be included as an 'extra' if desired;
- There is no methodology for assessing residual stresses. The joints are assumed to have been "stress relieved" or "temperature conditioned", or whatever other process is required by the relevant design code. TJs without any form of heat treatment would not be covered by R5V6. The procedure is remiss in not making this explicit.

Qu.: What are the limitations/restrictions on applying R5V6?

- The procedure addresses pipe circumferential butt welds only.
- The procedure can be applied only to those joints for which there is adequate cross-weld rupture data.
- The procedure applies only for elastic cycling. TJs which experience cyclic plasticity cannot be assessed using R5V6 at present.
- The procedure does not currently mention PWHT. However the understanding is that it would not apply to non-stress-relieved Inconel filled welds of the type fitted initially in HYA/HRA between the s/h outlet headers and the main steam pipework. It is assumed that the procedure is applicable to all welds which have received an approved PWHT or 'temperature conditioning'.

References

[1] M.C.Coleman, "TGN046 – Guidelines for the Inspection of Transition Joints on AGR Plant", BEG/SPEC/ENG/TGN/046, May 2010.

[2] S.J.Heath, J.C.P.Garrett, "Review of Background Information Relevant to Guidelines for Inspection of Transition Welds in AGRs", TD/SEB/MEM/1132/93, April 1993.

[3] "Dissimilar-Weld Failure Analysis and Development Program: Volume 1 Executive Summary", EPRI CS-4252, Project 1874-1 final report, November 1985.

[4] Ray Nicholson TPRD/M/1311/R83, June 1983 – It would be worth getting hold of this. It is listed on CDMS Records but not attached, so it will be on microfiche.

Fatigue in R5V2/3

I originally included this material in these notes but I now consider them out of place. They are retained here only for my personal record and will not be part of the T73S06 series of tutorials.

Qu.: Does R5 Assess Gross Fatigue Failure or Fatigue Crack Initiation?

- R5V2/3 assesses creep-fatigue crack initiation, and hence the fatigue part necessarily relates to crack initiation.
- R5V6: The fatigue assessment methodology for transition joints in R5V6 makes no mention of, or allowance for, crack initiation and must be regarded as a gross fatigue failure assessment.
- R5V7 of course does not assess fatigue (because it covers steady loading only).

Qu.: But doesn't R5V2/3 use fatigue endurance data?

Yes.

The input data used in R5V2/3 is from fatigue tests taken nominally to 'failure'. However, 'failure' in fatigue tests does not always mean that the specimen breaks into two parts. Most often fatigue tests are conducted in uniaxial strain control, and 'failure' is defined as the maximum load decreasing by 25%. However, this is taken to be equivalent to the specimen breaking if the test had been carried out at constant stress range.

Qu.: So how does R5V2/3 assess fatigue crack initiation?

An adjustment is carried out to the endurance data according to the depth of crack, a_0, which the User wishes to regard as "just initiated". This adjustment is specified in R5V2/3 Appendix A10, Section A10.1, and is summarised here:-

Crack Initiation Size Adjustment to Fatigue Endurance

Fatigue endurance data is obtained from specimens typically of diameter 6-10mm. The 'failure' of such a specimen is taken to consist of,

(a) a number of cycles, N_i, to initiate a very shallow crack of depth 0.02mm, and then,

(b) a further N_g cycles to 'fail' the specimen.

Suppose the endurance of the specimen is N_l cycles at the strain range of interest. Then $N_l = N_i + N_g$. The number of cycles to initiate a 0.02mm crack is taken to be the same in your structure as in the test specimens, and is given explicitly in terms of the total endurance, N_l (e.g., from R66 data) by,

$$\ln N_i = \ln N_l - 8.06 N_l^{-0.28} \qquad (1)$$

and hence N_g is found as $N_l - N_i$.

If your chosen a_0 is less than 0.02mm, then the initiation-endurance to use in the assessment is just N_i. In general, though, your chosen a_0 will be greater than 0.02mm, perhaps 0.2mm or some suitable percentage of the wall thickness. In this case some allowance for crack growth from 0.02mm to a_0 is required – an extra N'_g cycles. This is given by $N'_g = MN_g$ where the fraction M is, For $a_0 < 0.2mm$ $$M = \frac{a_0 - 0.02}{0.2\ln(a_l / 0.2) + 0.18} \tag{2a}$$

For $a_0 > 0.2mm$ $$M = \frac{0.2\ln(a_0 / 0.2) + 0.18}{0.2\ln(a_l / 0.2) + 0.18} \tag{2b}$$

where a_l is the diameter of the test specimen (probably 6mm-10mm). The initiation-endurance to use in the assessment is then $N_0 = N_i + N'_g$, and the fatigue damage per cycle is $\Delta D_f = 1/N_0$.

Qu.: Does fatigue damage depend upon stress triaxiality?

Yes.

When the assessment is close to uniaxial, the Mises equivalent strain range can be used to find the endurance, N_l, which feeds into the procedure stated above, i.e., Equs.(1-2).

If the assessment is shear dominated this procedure can still be used but will be conservative. For reduced conservatism use the multiaxial procedure.

Conversely, if there is significant biaxial or triaxial tension, the uniaxial methodology using the Mises strain range may be non-conservative and the multiaxial route should be followed.

Qu.: What is the multiaxial fatigue methodology?

The multiaxial procedure consists of replacing the Mises strain range as follows,

- Use the Tresca strain range to find N_l in order to find N_i from (1);
- Use the Rankine strain range to find N_l in order to find N_g and hence N'_g from Equs.(1-2);
- Set N_0 to the sum of N_i and N'_g from the above two steps.

The details of how to calculate the Tresca and Rankine strain ranges are given in R5V2/3 Appendix A10, Section A10.2.1. The advice covers only surface assessment points. The key steps are,

- First find the elastic-plastic Mises strain range from the hysteresis cycle construction of R5V2/3 (to be covered in later sessions);
- Enter the cyclic stress-strain curve for your material at the above strain range to find the secant modulus, E_s (i.e., the ratio of stress to strain at that point);
- Define the adjusted Poisson's ratio as $\overline{\nu} = \nu \frac{E_s}{E} + 0.5\left(1 - \frac{E_s}{E}\right)$;
- Identify the elastic stress ranges in the plane of the surface point in question, $\Delta\sigma_1$ and $\Delta\sigma_2$;
- Use R5V2/3 Equs.(A10.7-9) to find the elastic-plastic Tresca and Rankine strain ranges from the elastic stress ranges.

Qu.: Is the multiaxial methodology commonly employed?

No.

Perhaps it should be, but it isn't. There is an excuse, however. Generally fatigue damage is small. In which case it will remain small if the refinements of the multiaxial route were followed.

Qu.: How are different cycle types assessed?

Generally Miner's Law is used, i.e., the damage is given by $D_f = \sum_j \frac{n_j}{N_{0j}}$.

However, in truth, the damaging effect of a mix of different cycle types can depend upon their order. R5 suggests that order effects might result in the fatigue damage differing from that calculated by a factor of up to ~2. Some advice on this is offered in R5V2/3 Section A10.3. However, so long as D_f is small this will not be of great concern.

Qu.: What is the Fatigue Strength Reduction Factor (FSRF)?

The above procedure applies to smooth parent features. But weldments are more susceptible to fatigue damage. This is due to a combination of factors.

Firstly the geometry of weld can exacerbate fatigue, especially if the weld cap is still present. The stress concentration at a weld toe feature will generally have a very marked effect. This is where a crack will initiate.

Secondly, the weld material can differ from the parent, so that there is also a material effect.

Both effects are included in the FSRF, which is obtained by testing welded specimens or features. The FSRF is defined as the ratio of the plain parent strain range to the welded feature strain range giving the same endurance.

The recommended values for the FSRF for various weld types are given in R5V2/3 Appendix A4, Tables A4.1-3.

Qu.: Is the FSRF used to factor the stress range or the strain range?

The FSRF is a factor on strain range, *not* on stress or cycles.