

Tutorial Session 23: T73S06 (Creep Rupture / R5V6) – Creep Mechanisms

Relates to Knowledge & Skills items 1.1, 1.2, 1.4, 1.5, 1.6

Last Update: 13/7/14

What is creep? Deformation versus Failure; Mechanisms of creep; Mechanism maps; Failure versus crack initiation; Primary, secondary and tertiary creep versus the behaviour of real 300 series austenitics: examples of real creep strain v time data; Mechanisms of primary/secondary transition; Temperature dependence (Arrhenius); Mechanisms of tertiary creep; Example of isochronous data; R5 definitions of insignificant creep: is it really always insignificant? ‘Cold’ creep.

WARNING: Do not use these notes as a source of data for real assessments

Qu.: What is creep?

Almost everything about creep is complicated, uncertain and contentious, including this opening statement: creep is two distinct things,

- [1] Creep is the thermally activated, time dependent accumulation of irrecoverable strain;
- [2] Creep is the thermally activated, time dependent accumulation of damage potentially leading ultimately to failure.

In both cases it is understood that strain and damage might be accumulating at constant load – to distinguish creep from fatigue, which involves load cycling.

Generally these two things, creep strain and creep damage, are taken to be correlated, but this is more debatable than is often acknowledged (of which more below).

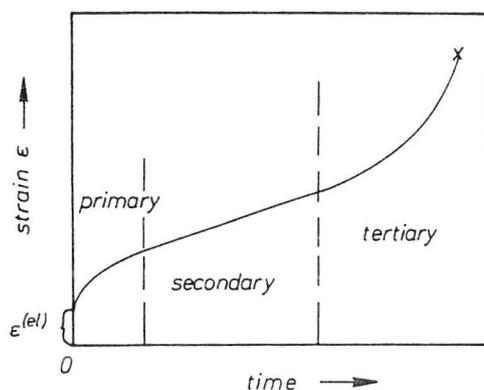
By ‘creep’ we shall mean *thermally activated* time dependent processes, and hence we are considering high temperatures (where ‘high’ is defined below). There is a phenomenon often loosely called ‘cold creep’, but this is actually a misnomer for time dependent plasticity (also discussed further below).

Qu.: Why is creep complicated?

At high temperatures there is a lot of thermal energy available. Consequently many mechanisms can be active which would not be active at lower temperatures. So, creep is complicated because there are many interacting processes happening at once.

Qu.: What do creep strain versus time curves look like?

Figure 1: *The archetypal primary-secondary-tertiary creep strain versus time curves*



For a given material there will be sets of such curves,

- (a) set of curves for different stresses (or loads) at the same temperature, or,
- (b) set of curves for different temperatures at the same stress (or load).

Creep data can be expressed using either engineering stress or true stress and using either engineering strain or true strain. Creep curves are generated from a single specimen. Since creep rigs are most often dead weight loaded, the load, and hence the engineering stress, is constant, but the true stress will increase with time. The true stress is usually not known since the reduction of area is measured only at the end of the test. An estimate of true stress based on elongation (from extensometry) can be made, but is likely to be inaccurate as regards the behaviour of the region which ultimately fails (necking).

An important issue is the loading strain, i.e., the strain which occurs "instantaneously" on loading. For applied stresses above yield the plastic strain can be large. In some tests the plastic strain can considerably exceed the creep strain at failure. In such cases the specimen never creeps at the engineering stress since the true stress is substantially greater even at time zero. It is particularly important in such cases to analyse the data using the true stress, for which the loading strain is required. Unfortunately the loading strain was not recorded in many older tests! (Note also that even the plastic strain does not really occur instantaneously).

An indication of the magnitude of the correction to the creep strain which results from an elongation-based correction to stress is shown in Fig.2 (based on creep strain alone).

Figure 2: *Illustration of how the creep strain in a constant load test differs from that expected on the basis of constant stress*

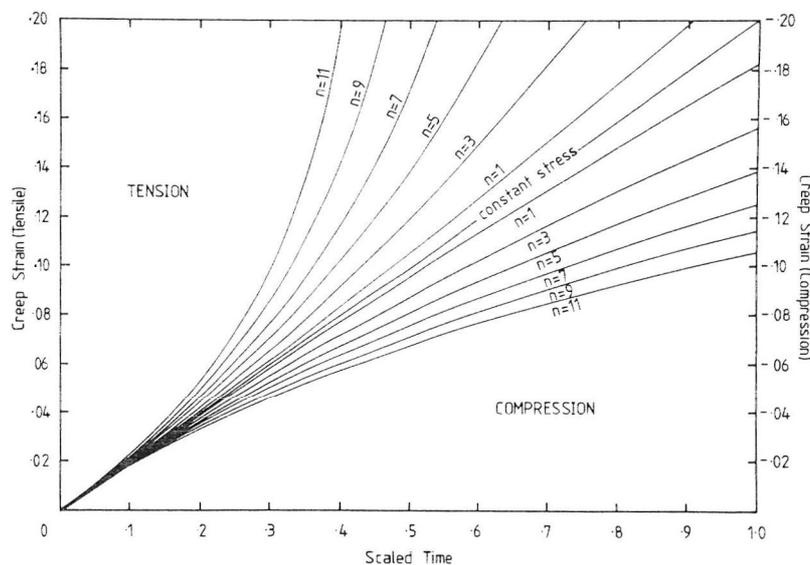
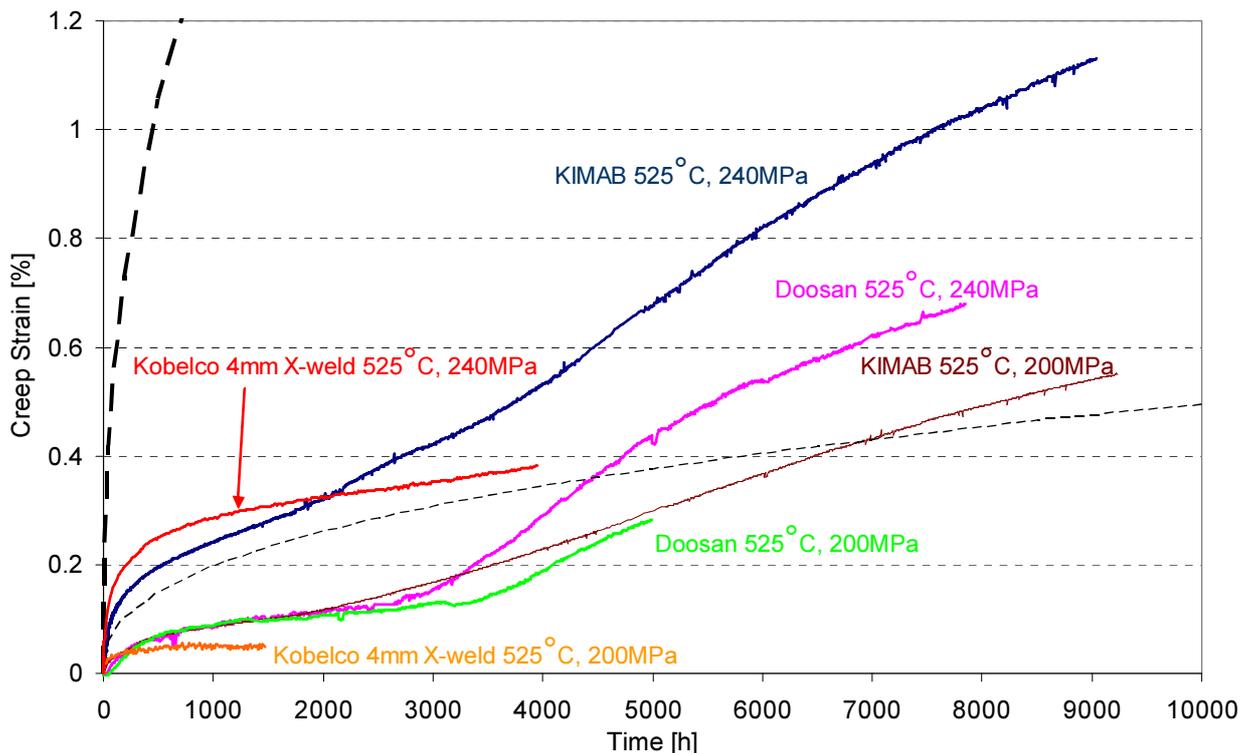


Figure 2.11 Constant load creep curves in tension and compression for a material which exhibits a linear variation of strain with time at constant stress. Under constant load conditions, the strain time behaviour departs increasingly from the linear constant stress line as the stress sensitivity, n , of the creep rate increases.

Qu.: Are creep curves always like Figure 1?

No. Sometimes they are. But reality is sometimes more complicated. Fig.3 is an example of 316H creep data showing “double-primary” behaviour.

Figure 3: Examples of “Double-Primary” Creep: *The double primary behaviour is found for virgin 316H parent, but not for welded specimens or material which has seen prior straining, e.g., in service or due to pre-conditioning. (The dashed black lines are RCC-MR curves, not experimental data). Note that the elastic-plastic loading strain has been subtracted from the data below. It is likely that none of the tests depicted have yet reached secondary creep.*



Qu.: What other ways of plotting deformation data are there?

Isochronous curves – plots of stress against total strain, each curve being at a fixed time. The total strain comprises elastic, plastic and creep strains. So, at sufficiently early times, this is just the elastic-plastic stress-strain curve. The creep strain at a given time is obtained as the difference between the isochronous curve and the zero-time curve. An example of isochronous curves is shown in Fig.4(b).

Such curves are intended to be obtained from constant stress tests, but care may be needed to determine if this is indeed the case (i.e., constant load test data may have been used, fitted to a creep strain expression and this fit used to derive the isochronous curves. Noting the difficulty of correcting constant load tests to true stress, this may not be the same thing). Also, the strain plotted could be engineering strain or true strain. The moral is that you need to dig into the detail to discover what any given creep curve or isochronous curve really means.

Figure 5: Example of Rupture Data: Stress against rupture time for a range of temperatures. This shows the 'Soviet model' fit from R66. (Now replaced in R66 by Sarah Spindler's improved fit).

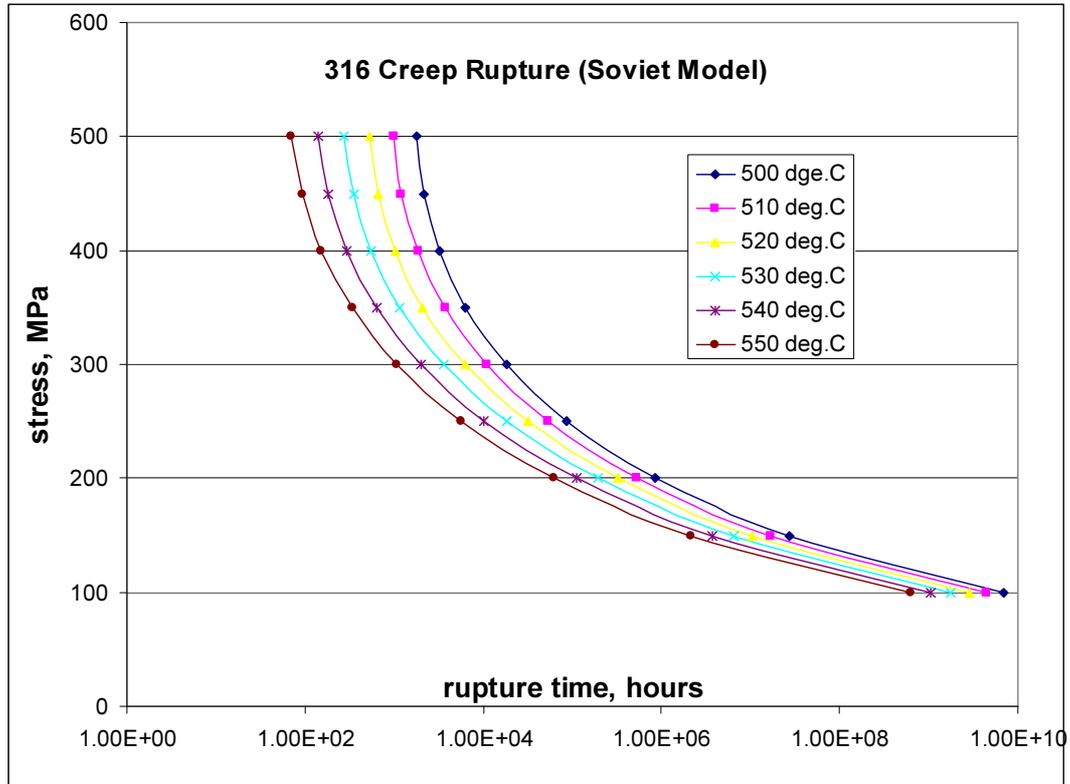


Figure 6: Example of rupture stress versus Larson-Miller parameter

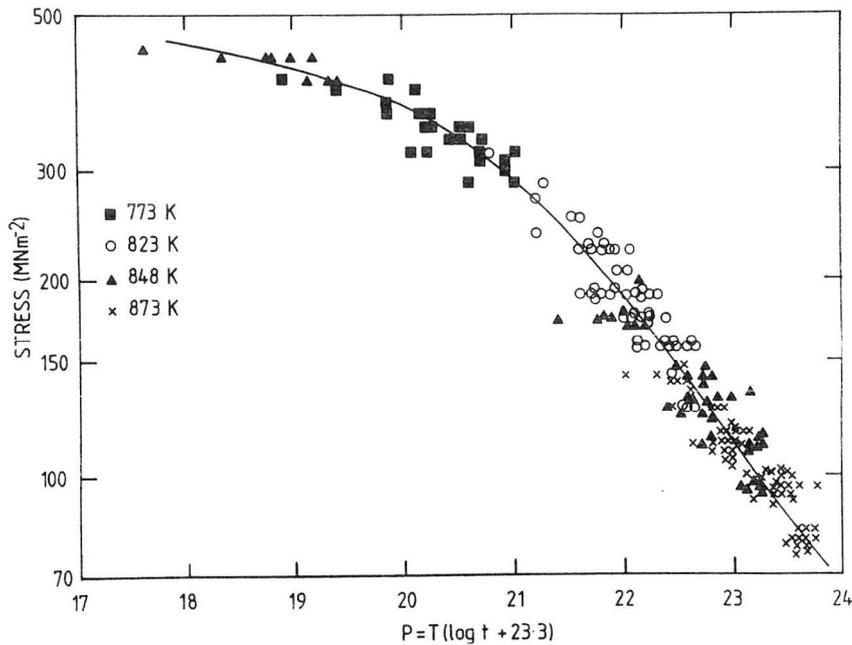


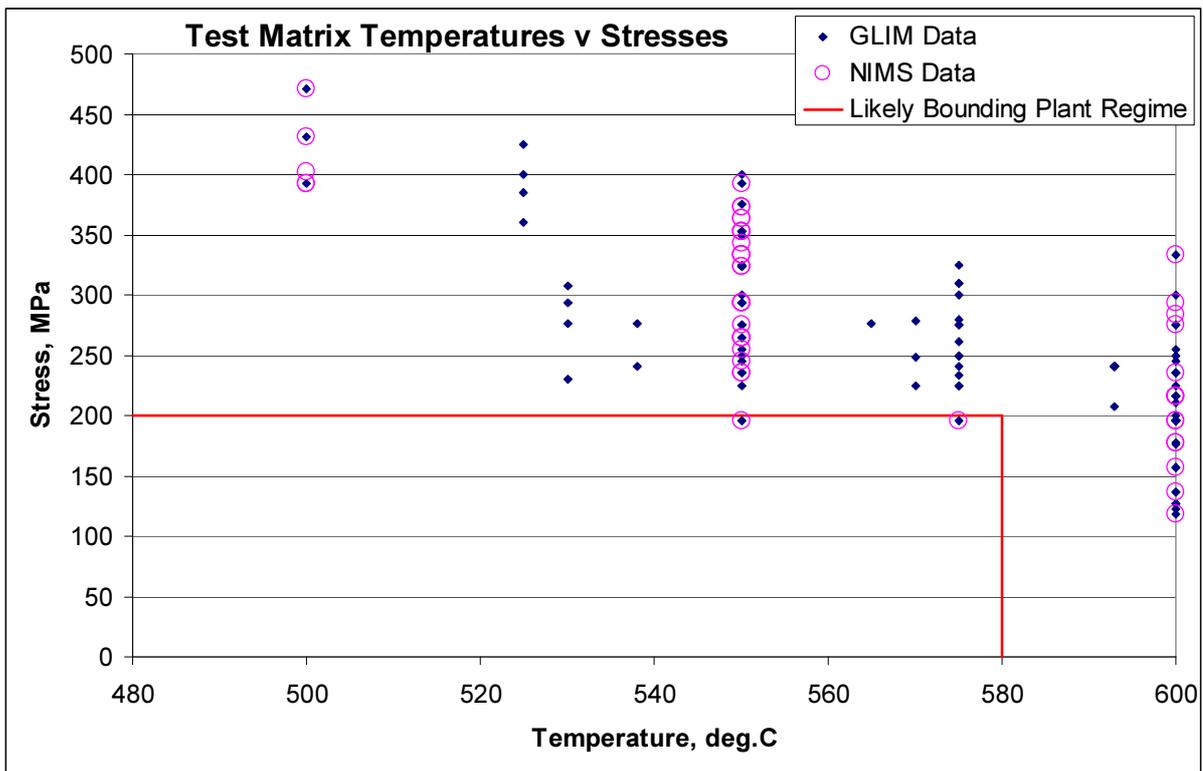
Figure 5.21 Stress rupture data for $\frac{1}{2}\text{Cr}\frac{1}{2}\text{Mo}\frac{1}{4}\text{V}$ steel showing that the results obtained under constant load testing conditions at temperatures from 723 to 873K can be superimposed onto a master curve using the parametric relationship suggested by Larson and Miller (332).

Qu.: What uncertainties are there in creep rupture data?

Very considerable uncertainties! These are due to (i)scatter - which is horribly large in creep, and, (ii)extrapolation - which is also a major problem in creep. An illustration of the extrapolation problem is given by Figures 7a-c for 316H stainless steel. For a fairly extensive database of creep tests, Figure 7a shows what combinations of stress and temperature were covered. It also shows what combinations of stress and temperature are likely to be relevant for superheater boilers (where many of our problems have been). There is virtually no overlap between the two!

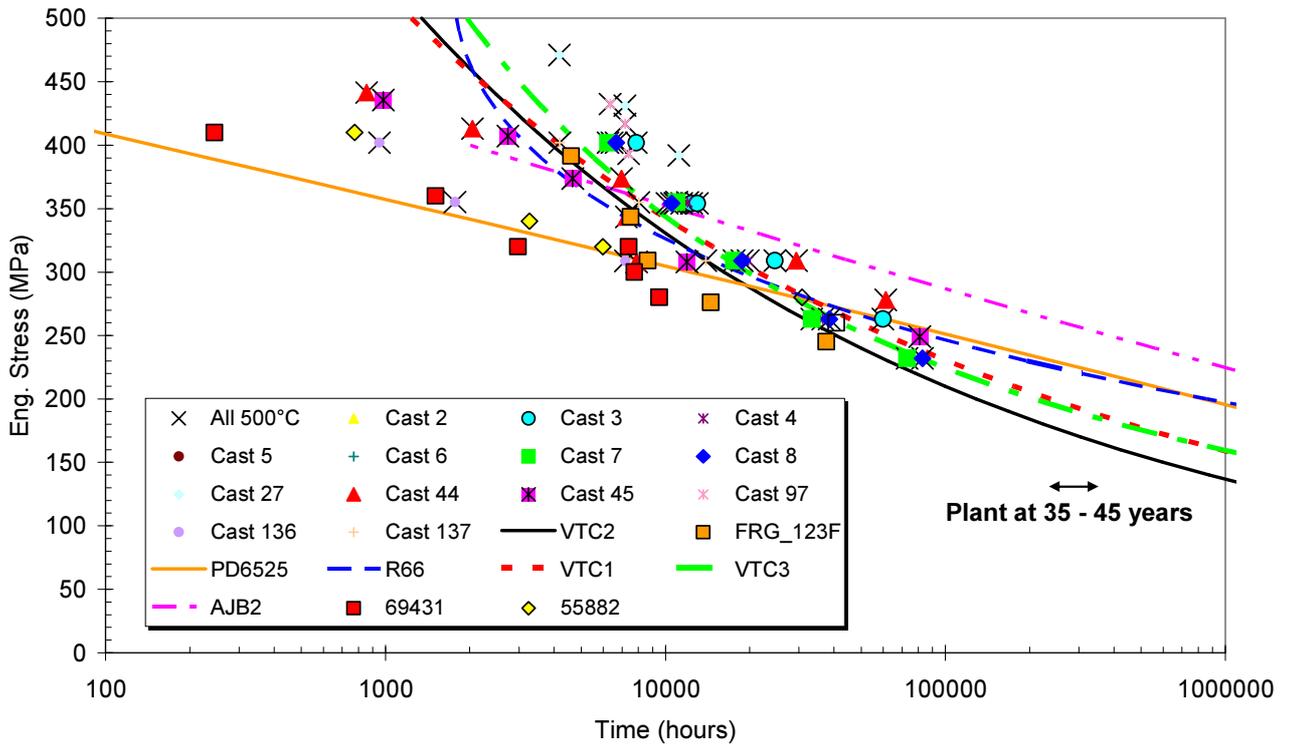
Figure 7b shows the rupture data on the usual log-time plot, together with a curve fit which can be used to extrapolate to the required plant conditions. This does not look too bad. But the same data on a linear-time plot, Figure 7c, reveals the problem more starkly. Like it or not, the extrapolation of the rupture data to the lower plant stresses and longer lifetimes is obviously extremely uncertain.

Figure 7a Combinations of Stress and Temperature covered by a Test Database compared with the Combinations Relevant to Superheater Boilers* (316ss)

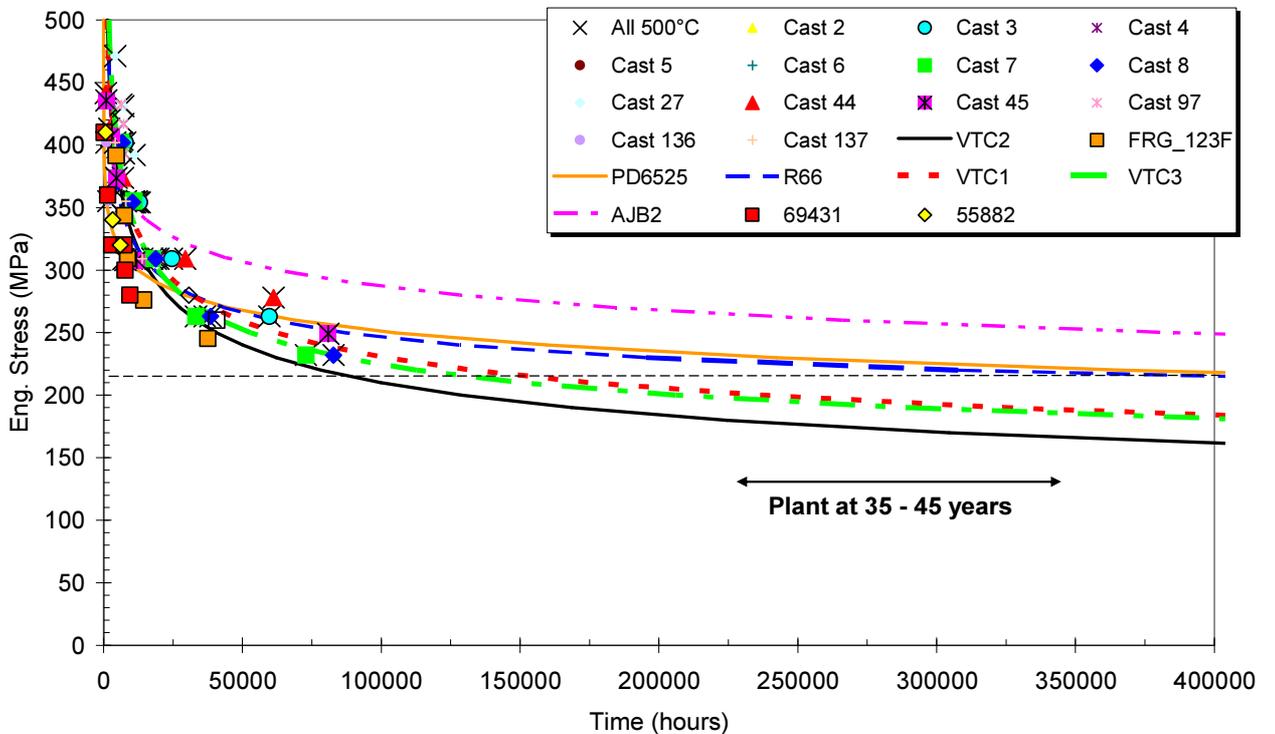


* Reheater boilers might be hotter, and components within the T2 parts of the reactor and in contact with the coolant gas will certainly be hotter (up to 650°C), so some of the extreme right-hand data might just be appropriate for these plant areas.

**Figure 7b 316H Rupture Data versus Log-Time:
Extrapolation to Plant Looks Reasonable**



**Figure 7c 316H Rupture Data versus Linear-Time:
Extrapolation to Plant Looks Rather Arbitrary**



Qu.: What other creep failure data is there?

The other key piece of information from a failed creep test is the strain at failure. The creep strain at failure is the creep ductility. (At least, it is one definition of creep ductility). The creep strain is the total strain minus the initial loading strain, the latter being the sum of the elastic and plastic strains. Beware, though - some sources may quote the total strain at failure, including the plastic strain, as being the creep ductility. Even R66 used to do so in the early 90s for some materials. And the “stress dependent ductility” model, to be incorporated into R5V2/3 Issue 3 at Revision 2 in 2014, is also based on total inelastic strain, plasticity plus creep.

Materials can have ductility troughs in certain temperature ranges, illustrated for 316/316H below...(creep strain only)...

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

So - the ductility is lower at the lower temperatures, then? Is it? Are you sure?

Things are not so simple...

- In order to get test results in a reasonable time, tests at the lower temperatures tend to use higher stresses (Figure 8c). So the truth might be that creep ductilities are lower at higher stress.
- But higher stresses correspond to larger plastic loading strains - so the truth might be that creep ductility is smaller for higher plastic strains. If so, that might imply that the total (plastic plus creep) strain at failure is closer to being a constant after all.
- Or alternatively, it might be that the lower temperature data corresponds to lower strain rates - and creep ductility is often considered to depend upon strain rate, showing a 'lower shelf' behaviour at sufficiently small strain rates, e.g., see Figs 8a and 8b below...

Fig.8a CMV Creep Ductility 475-525°C (R66 Rev.008)

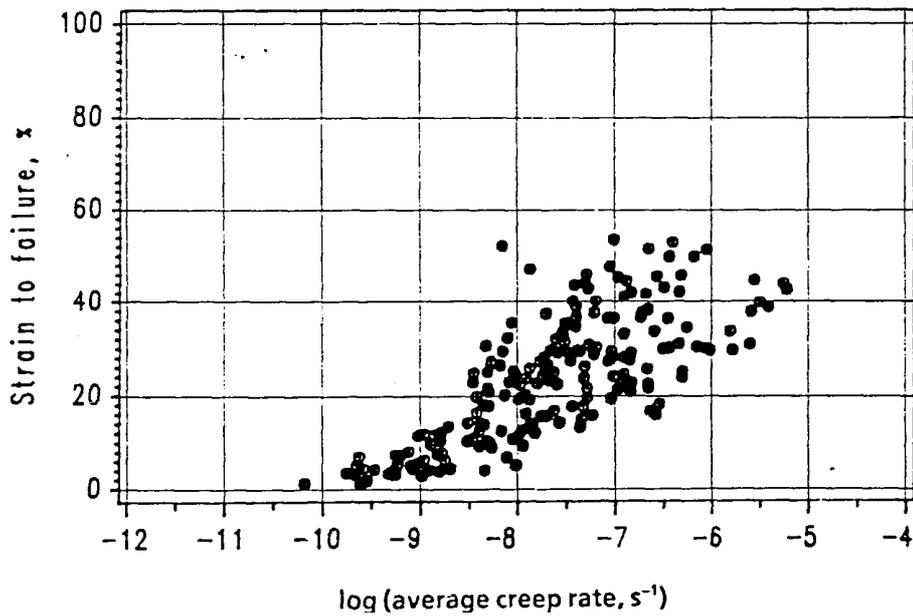


Fig.8b Ex-HYA S/H Header 316H 525-575°C Creep Ductility (R66 Rev.008)

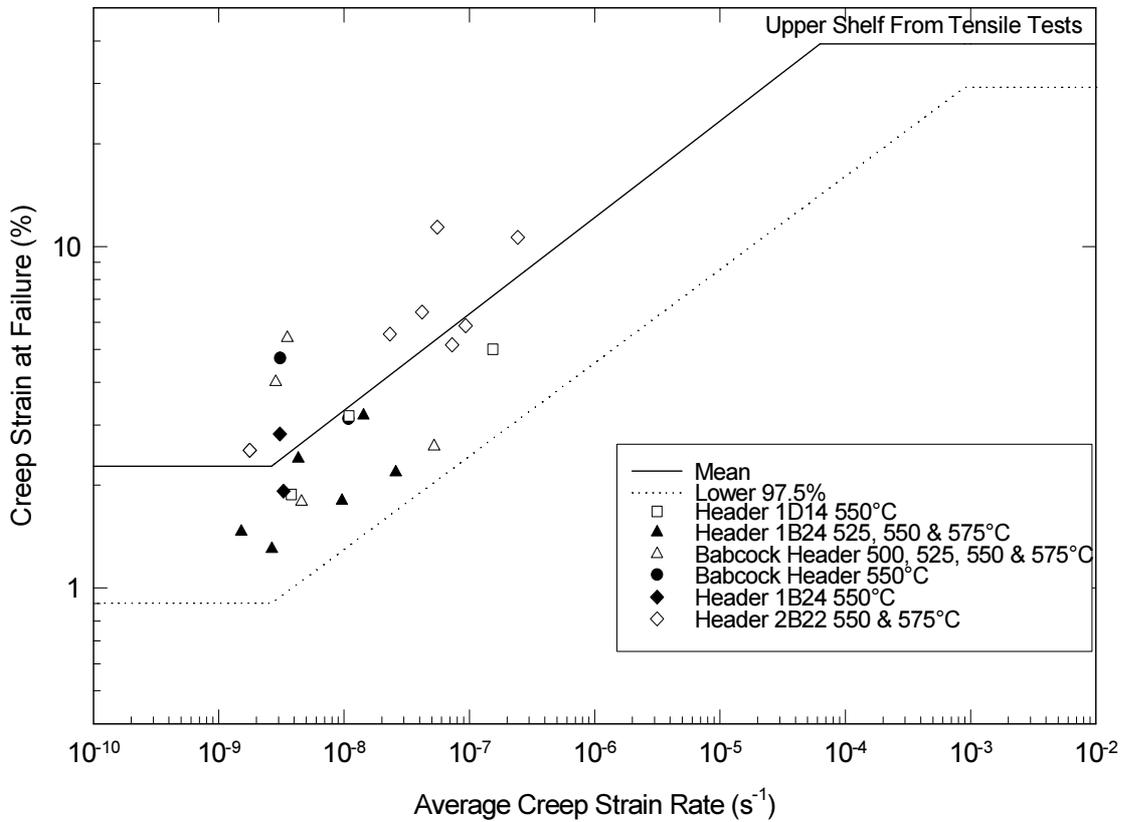
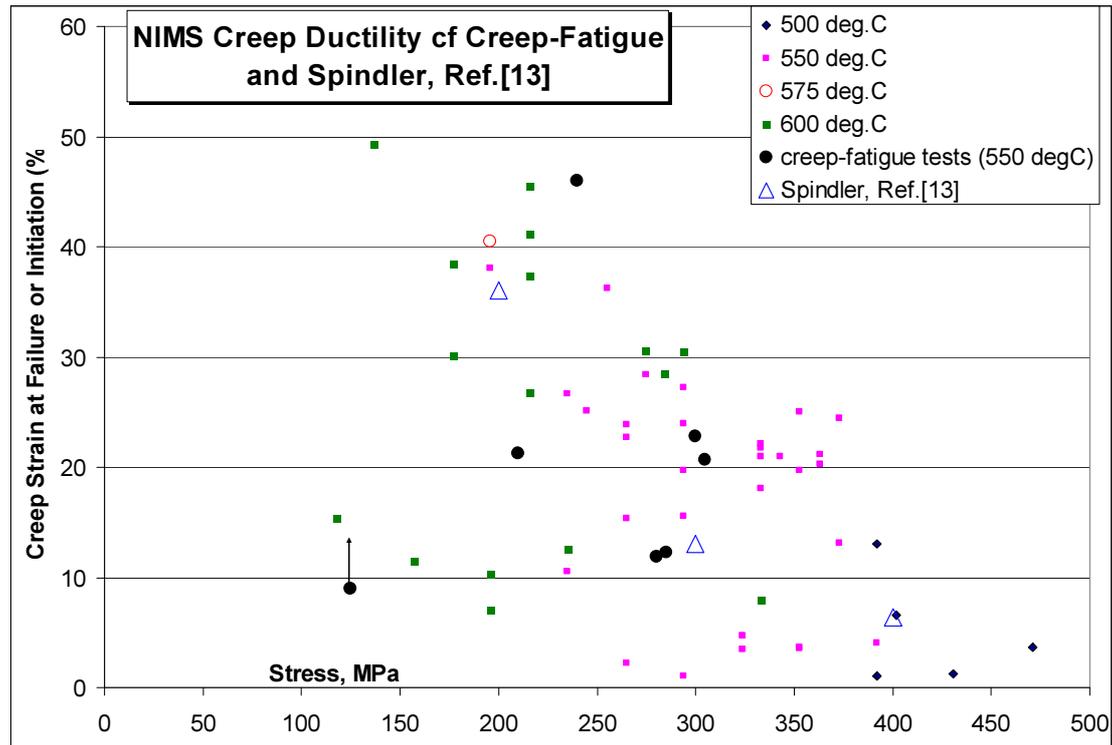


Figure 8c Large Scatter in 316H Creep Ductility and its Dependence on Stress or Temperature (but not clear which!)



Qu.: What uncertainties are there in creep ductility data?

Just as for rupture data, there is huge scatter (Figure 8c) - ductilities range from ~1% to ~50%. And there is again the same issue as regards extrapolation (Figure 7a again applies).

The moral is,

- Be aware of the great uncertainty in materials data, especially in creep. If you just take curve-fit recommendations from R66 and never look at the raw data you will not be aware of how vulnerable we are on this issue;
- The extrapolations to plant conditions might be more credible if the curve-fits were constrained by mechanistic considerations - which is generally not currently the case but may be where we going;
- Arbitrary 'bounds', e.g., a 95% confidence limit (say ~2.6% for Figure 8c) are just that - arbitrary. So what do deterministic assessments mean?
- Probabilistic assessments are strongly motivated - but we need to be able to do the assessment deterministically first.

Qu.: Consistency of creep data?

In principle, if you are given fits to stress-rupture data and fits to strain-time data then you should be able to deduce the creep ductility at a given stress and temperature (by finding the time to failure from the rupture fit, and entering the strain-time curve at this time). However, if ductility data is fitted separately, substantial inconsistencies can arise simply due to,

- (i) the way in which the fitting is carried out on highly scattered data, and,
- (ii) ambiguities in the definition of “creep ductility”.

The first of these can be avoided by fitting techniques which constrain each quantity together. There are ways of doing this, e.g., Baker-Cane, but Baker-Cane is probably not good enough. The HiTBASS programme is attempting to address issues such as this. The ideal is to develop mechanistically inspired models which address both deformation and the time and strain at failure. But it isn't easy.

Qu.: What about creep under multi-axial stressing?

All the examples of data illustrated above are for simple uniaxial tension. Creep under multi-axial conditions is discussed in a later session. The differences are very important.

Qu.: What is meant by “hardening” in creep?

There is a risk of confusion over “hardening” in creep because the word is used in at least two different senses – and both differ from its meaning in plasticity. The first usage simply refers to creep rates. During primary creep the material is described as ‘hardening’, simply meaning that the creep rate is reducing at constant stress. The converse is a ‘creep-soft’ material which means a material which will deform in creep relatively quickly.

The second usage is in the phrase “hardening law”. This refers to the procedure being used to take account of previous creep when calculating current creep rates. Fig.4 illustrates one hardening law, namely “strain hardening”. Hardening laws will be discussed in more detail in a later session. A hardening law is required only if conditions are changed (e.g., load is changed, or temperature is changed).

Qu.: What about creep under displacement control (relaxation)?

Tests under displacement control are sometimes carried out. However, given the above creep data, and a hardening law, you should be able to calculate the outcome of a test under displacement control, or secondary loading. So such tests can be used as validation of our understanding.

Unfortunately, that “should be able to calculate the outcome” has generally not been realised in practice. For example, relaxation data often lies *between* predictions based on integrating forward creep with and without the primary creep period. Similarly there is evidence that changing temperature affects creep rates in a manner which is intermediate between that predicted by time hardening and strain hardening.

Note the relevance of the hardening law in relaxation. Conditions inevitably change under displacement control, since the stress relaxes. Consequently relaxation cannot be calculated from a forward creep law unless some hardening behaviour is assumed.

Qu.: What about data for crack initiation?

It is perfectly possible to carry out tests which are designed to identify the conditions of time, stress and temperature leading to crack initiation. This is most commonly done for creep-fatigue, in which the load is periodically cycled as well as sustained at high temperature. However, tests of this type are not generally used to provide ‘materials data’. Instead, crack initiation is assessed (e.g. to R5 V2/3) using creep strain rate and ductility as the input materials data. Tests leading to crack initiation can then be used as a validation of these assessments.

Qu.: At what temperature does creep become significant?

Time, temperature and stress are all important to whether creep is significant. Thus, the temperature at which creep is insignificant for a given material must be stated with reference to a specified time period and a stress level. Generally the time period will be the plant design life (or extended life).

The R5 definition of insignificant creep is given in R5 V2/3 Appendix A1, Section A1.6, and is that,

- EITHER, relaxation from an initial stress of 1.35Sy, at constant total strain, produces less than 20% relaxation of stress;
- OR, the creep strain accumulated at a constant stress of 1.25Sy is less than 0.03%;

where Sy is the lower bound 0.2% proof stress. The resulting temperature, for the stated time period, is known as LOIC (Limit of Insignificant Creep).

R5 Table A1.1 gives numerical values. Choosing a time of 270khrs (or up to just above 300khrs), i.e., crudely a 40 year life at realistic average load factors, gives LOIC values of *roughly*,

½CMV	420°C
1Cr½Mo	390°C
2¼Cr1Mo	~345°C
9Cr1Mo	380°C
304ss	~445°C
Alloy 800	476°C
316ss	470°C

(Don't use these in assessments – look in R5 or work out your own)

Note that LOIC does not simply increase monotonically with the % alloying elements.

Figure 5.3 of R5 V2/3, copied below as Fig.9, shows how LOIC varies with the time period assumed.

In an R5 assessment, creep can be ignored below LOIC. However, you should not interpret this to mean that creep effects are entirely absent below LOIC. Many creep tests have been carried out under the HPB/HNB BLIP program on 316H at 470°C (the LOIC for a 40 year life). Both creep strain rates and creep crack growth rates are sometimes (not always) measurable – even over periods of less than a year or so, let alone 40 years. Having said that, the strains in question are usually only just within the measurable range (and in some cases may just be experimental error) – and there is no prospect of failure, of course. I have even seen measurable creep crack growth rates in 316(N) reported at 400°C!

Figure 9: LOIC Curves from R5 (Issue 3)

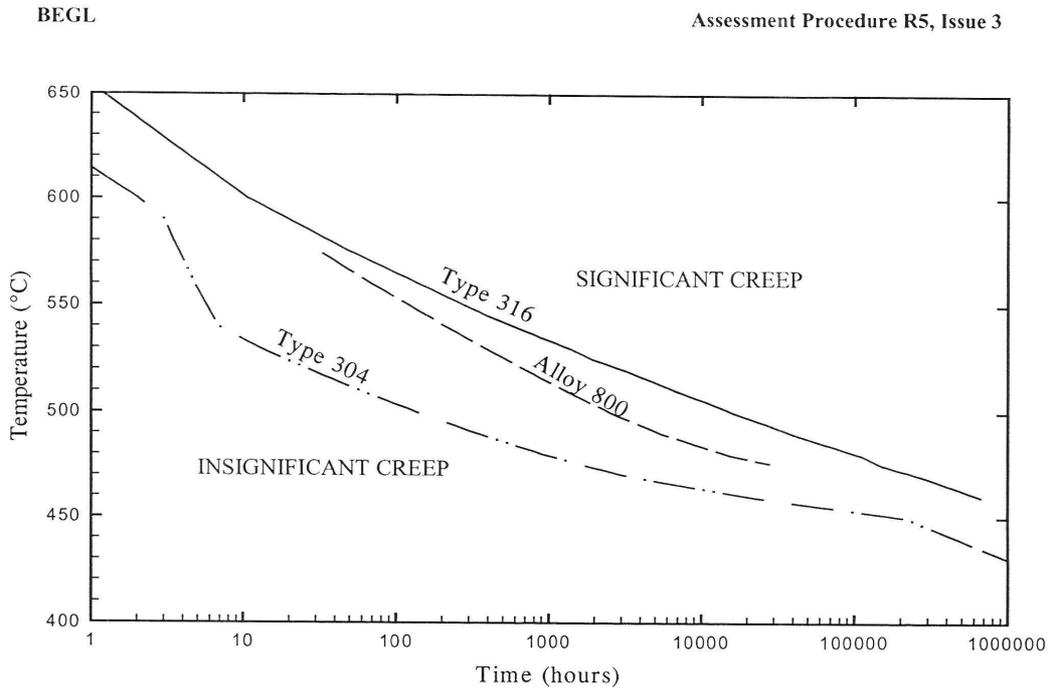


Figure 5.3(a) Insignificant creep curves for austenitic steels

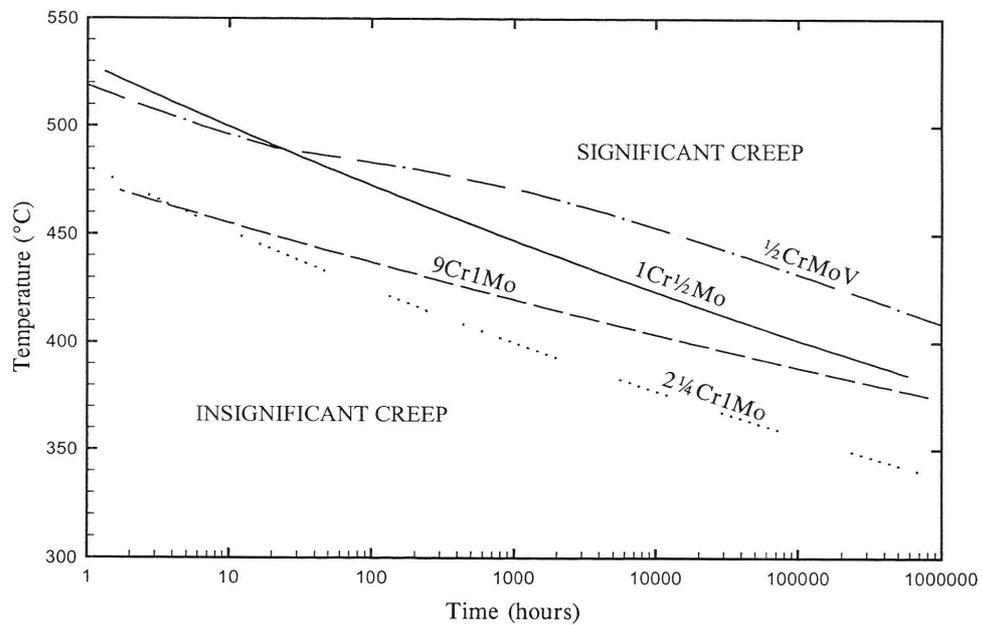


Figure 5.3(b) Insignificant creep curves for ferritic steels

Qu.: Doesn't 'cold creep' undermine LOIC?

Time dependent strain accumulation can occur well below LOIC, and even at room temperature, in some materials – if the stress is high enough. But this does not contradict LOIC. The loose term 'cold creep' really means time dependent plasticity, or anelasticity. In terms of macroscopic strain development, the effects of creep and anelasticity may be indistinguishable. They differ, though, as regards the mechanisms involved. Because the thermally activated mechanisms which lead to creep damage are not active when 'cold', the anelasticity occurs without any attendant creep damage. However, low temperature time dependent strain accumulation can reduce the plastic collapse load. For this reason the effect is addressed in R6 (see R6 Chapter III, Section III.4, "Sustained Loading"). At the time of writing some salient features are,

- Time-dependent plasticity occurs only beyond general yield, and hence need not be considered when $L_r < 1$;
- For ferritic steels, time-dependent plasticity is negligible for $L_r < 0.9L_r^{MAX}$;
- For austenitic steels, time-dependent plasticity may be neglected above 200°C.

Consult R6 for the definitive position. However, in no case does time dependent plasticity contribute to creep damage – and hence it is irrelevant to creep assessments.

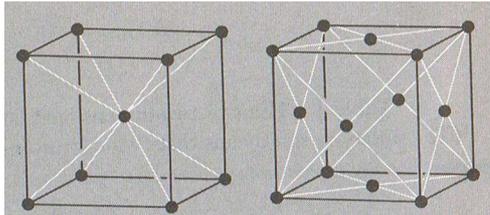
Qu.: What are the mechanisms of creep deformation (strain)?

There follows a very rapid review of basic material science. None of this is really required for SQEP in the T73 sub-roles, but a degree of appreciation of these mechanisms helps explain why creep data is so messy and scattered.

Qu.: What do BCC and FCC mean, and which is ferritic/austenitic?

- BCC = body-centred cubic crystal lattice structure (Fig.10)
- FCC = face-centred cubic crystal lattice structure (Fig.10)

Figure 10: BCC and FCC



- Alpha-iron (ferrite) = body-centred cubic
- Gamma-iron (austenite) = face-centred cubic
- FCC is close-packed, attaining the theoretical maximum packing fraction of 74%. Consider planes of hexagonally close packed atoms. Referring to the based plane as A, the possible positions of the next plane are B and C (see Fig.11). The FCC structure is equivalent to the stacking ABCABCABC... This is not immediately obvious, but the close-packed planes are those normal to the principal diagonals of the FCC cube, i.e., the $[111]$ planes.
- Hexagonal close packing is ABABAB... (Fig.11a). It also attains the maximum packing fraction of 74%, but is not the predominant structure in the steels we normally encounter.
- BCC is not close packed. It has a packing fraction of 68%. There are no close packed planes in the BCC structure.

Figure 11a: FCC as a close-packed structure ABCABCABC...

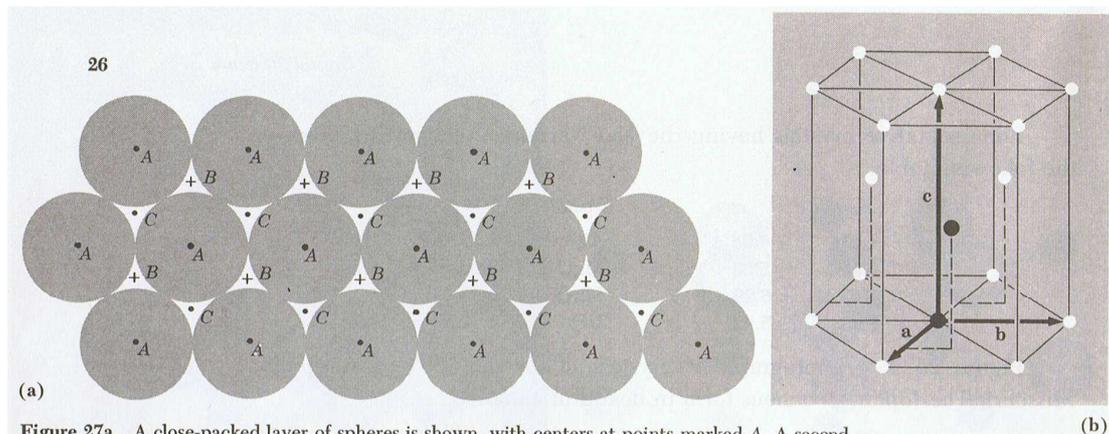


Figure 27a A close-packed layer of spheres is shown, with centers at points marked A. A second and identical layer of spheres can be placed over this, with centers over the points marked B (or, equivalently, over the points marked C). If the second layer goes in over B, there are two nonequivalent choices for a third layer. It can go in over A or over C. If it goes in over A the sequence is ABABAB... and the structure is hexagonal close-packed. If the third layer goes in over C the sequence is ABCABCABC... and the structure is face-centered cubic; the plane is a $[111]$ plane.

Figure 11b: Iron-Carbon Phase Diagram

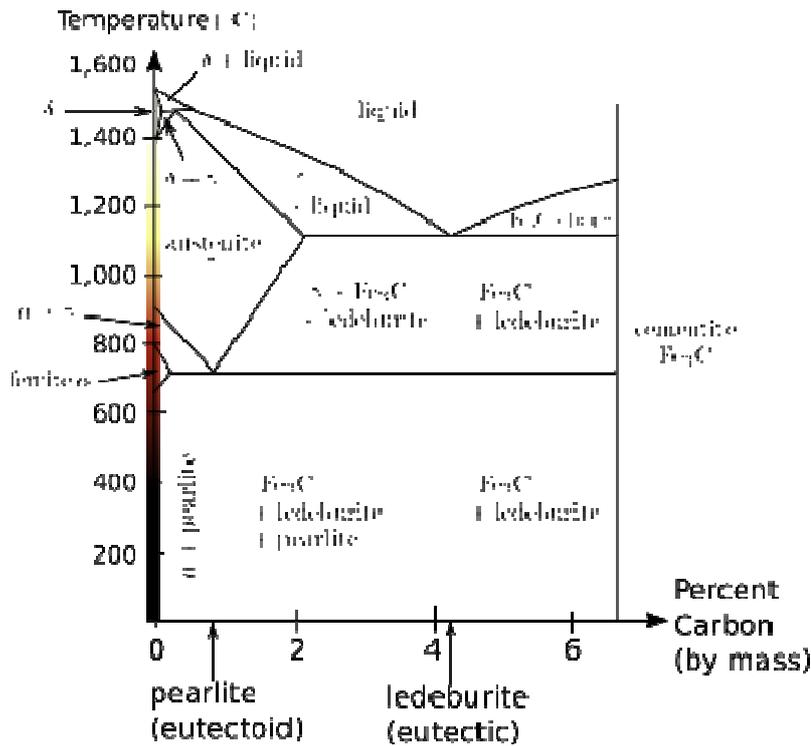
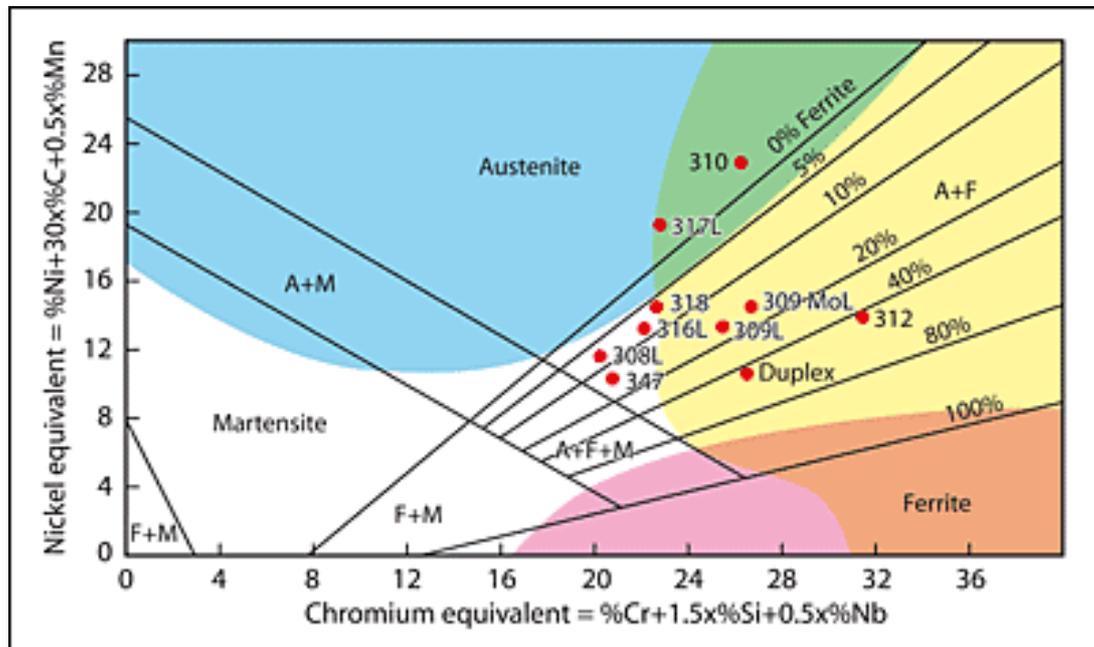


Figure 11c: (One Version of the) Schaeffler Diagram



Qu.: What sorts of defects occur in the crystal lattice?

- Edge dislocations
- Screw dislocations
- Vacancies
- Interstitials
- Impurity or alloy atoms
- Precipitate particles (e.g., carbides)
- Jogs and kinks on dislocation lines
- Dislocation pile-ups/networks
- Grain boundaries
- Twins

...and all these defects interact!

NB:

- Jog = step in dislocation moving it out of the slip plane
- Kink = step in dislocation within the slip plane

Qu.: What are “edge” and “screw” dislocations?

An edge dislocation can be thought of as an extra half-plane of atoms inserted into the lattice (or a half-plane removed), see Figs.12,13. It has a unique slip plane.

A screw dislocation is defined by its displacement mismatch (Burgers vector) which is parallel to the dislocation line, see Fig.14. Thus a screw dislocation is rather like a crack under Mode III shear. However, there is no ‘crack’! And so there is no preferred slip plane.

An edge dislocation glides in a manner like a crack under Mode II shear, as illustrated by Fig. 15. This Figure also shows that both edge and screw dislocations can result in the same deformation.

Figure 12: *Illustration of an edge dislocation*

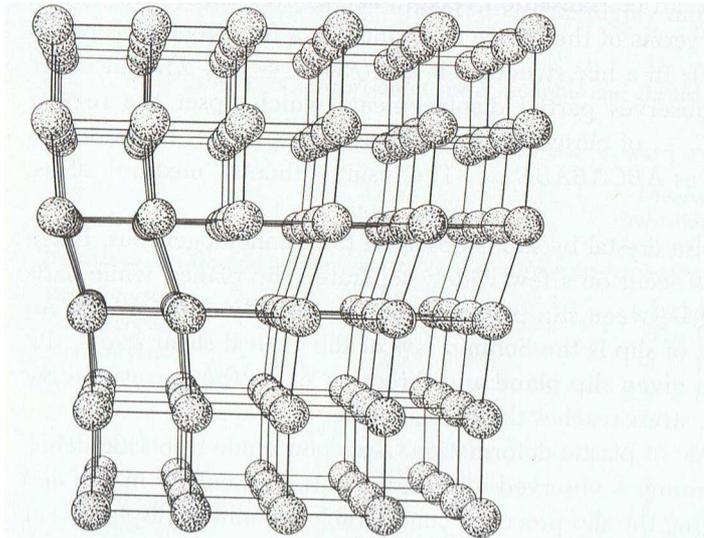


Figure 4 Structure of an edge dislocation. The deformation may be thought of as caused by inserting an extra plane of atoms on the upper half of the y axis. Atoms in the upper half-crystal are compressed by the insertion; those in the lower half are extended

Figure 13: *Bubble raft showing an edge dislocation (view at glancing angle)*

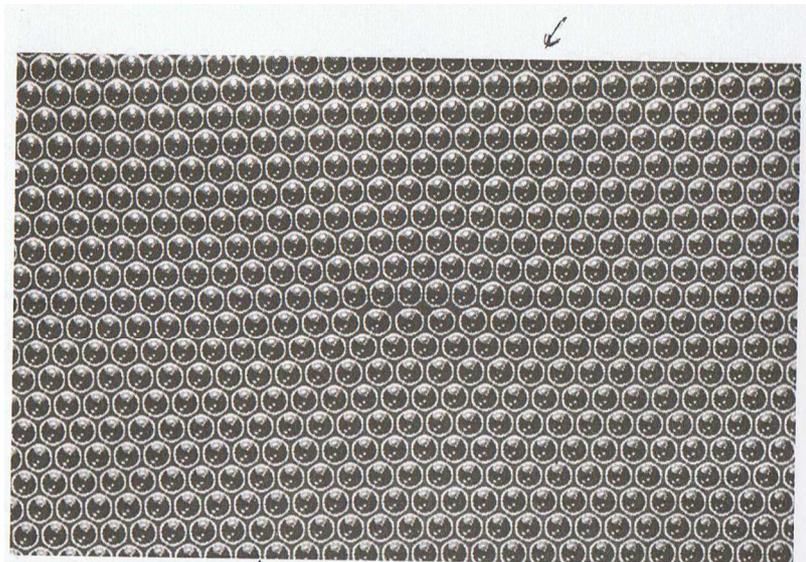
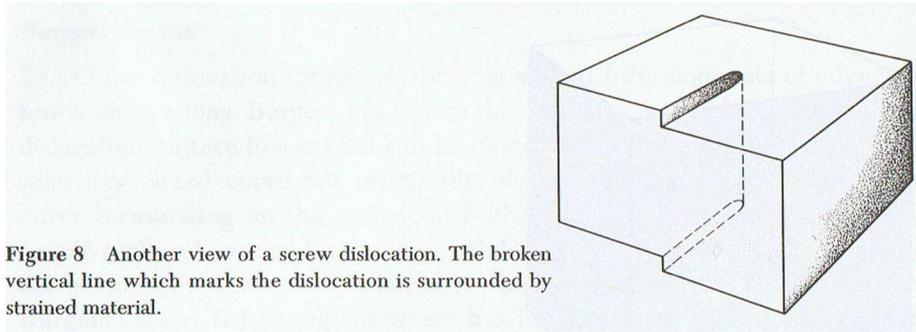
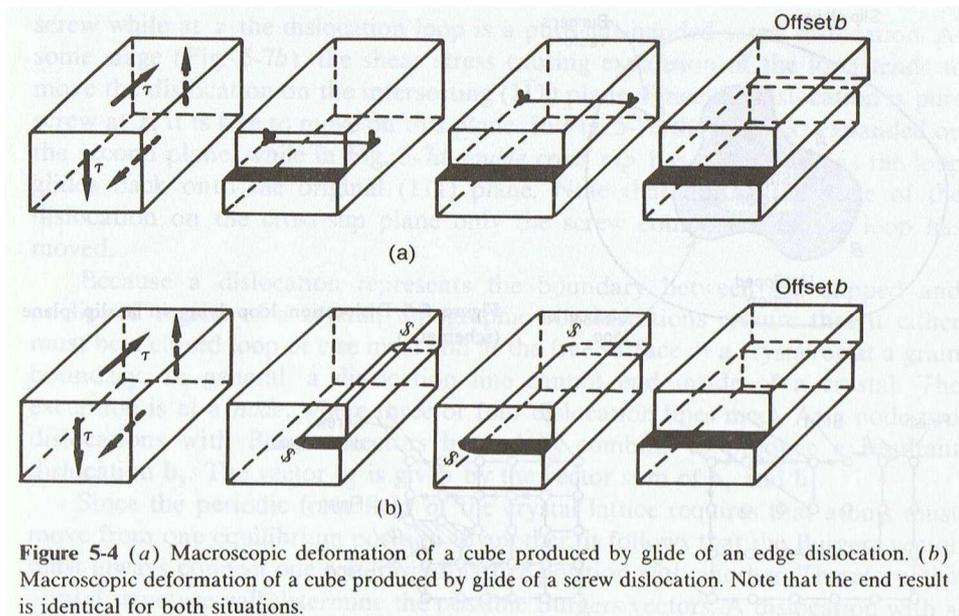


Figure 5 A dislocation in a two-dimensional bubble raft. The dislocation is most easily seen by turning the page by 30° in its plane and sighting at a low angle. (Courtesy of W. M. Lomer, after Bragg and Nye.)

Figure 14: *Illustration of a screw dislocation***Figure 8** Another view of a screw dislocation. The broken vertical line which marks the dislocation is surrounded by strained material.**Figure 15:** *Edge and screw dislocations can produce the same deformation***Figure 5-4** (a) Macroscopic deformation of a cube produced by glide of an edge dislocation. (b) Macroscopic deformation of a cube produced by glide of a screw dislocation. Note that the end result is identical for both situations.

Qu.: What is the influence of lattice defects on strength?

- Recall that theoretical strength of a perfect crystal with no dislocations is of the order $\sim G/2\pi$, which is huge. Nevertheless this strength can nearly be realised in thin whiskers grown virtually dislocation-free, thus demonstrating that dislocations are responsible for the lower strength of most materials.
- In contrast, in single crystals of very high purity, but containing dislocations, the yield strength can be very low. The purer the sample, the lower the strength. This demonstrates the importance of impurities (alloy atoms, inclusions, etc) in raising the strength of most structural materials. Note that the strength of very pure samples increases steeply (i.e., the material hardens rapidly) when the dislocations are used up, again confirming the weakening effect of free dislocations.

Qu.: What is meant by “activation energy” and “Arrhenius factor”?

Processes which are thermally activated proceed at a rate which is proportional to the so-called Arrhenius, or Boltzmann, factor $\exp\left\{-\frac{U}{kT}\right\}$, where U is the activation energy for the process, k is Boltzmann’s constant (1.38×10^{-23} J/K), and T is the absolute temperature (K). In statistical mechanics (thermodynamics), this factor is the fraction of “systems” (e.g., atoms) which will be found in a higher energy state, U , rather than at the datum zero energy, simply as a consequence of random thermal fluctuations.

Qu.: How many defects are present, e.g., vacancies?

- The energy required to create a vacancy is $\sim 1\text{eV}$. This is low enough for vacancies to be formed spontaneously by thermal vibrations of the atoms.
- Hence, at 500°C by using the Arrhenius factor it is found that a fraction $\sim 3 \times 10^{-7}$ of lattice sites are vacant, whereas at 1000°C , a fraction $\sim 10^{-4}$ of lattice sites are vacant – assuming thermal equilibrium.
- But rapid quenching from, say, solution heat treatment temperature (1050°C) can ‘freeze-in’ the higher vacancy fraction, i.e., $\sim 10^{-4}$.
- Vacancies are very important because, as we will see later, they are instrumental in creep deformation and also in creep damage and failure.
- But note that even the solution heat treatment vacancy fraction (10^{-4}) is still far too small to account for creep failures, i.e., it would correspond to a creep ductility of only 0.01%. Hence, creep failure must involve vacancy production.

Qu.: How many dislocations are there?

- The energy required to create a dislocation is much higher than for a vacancy ($\sim 8\text{eV}$ per atom plane). As a result, dislocations are not created thermally, in contrast to vacancies. Hence they must be produced by other means, principally by plastic deformation. All steels undergo severe deformation in their fabrication process, so dislocations are present from the start.
- Heat treatment will reduce the dislocation density. However, even annealed material typically has $10^3 - 10^6$ dislocations per mm^2 .
- Heavily cold worked material may have as many as $10^8 - 10^{10}$ dislocations per mm^2 .

Qu.: What mechanisms cause creep strain?

- Dislocation glide (slip)
- Grain boundary sliding
- Diffusion (of atoms and vacancies)

All three mechanisms might contribute to the materials and conditions of interest to us.

Qu.: What influences the strain rate due to dislocations?

- Glide will occur only on certain crystallographic planes and in certain directions – so the orientation of a grain with respect to the applied stress is relevant.

- Dislocations only glide so far before getting pinned on something, e.g.,
 - They get pinned on alloy atoms;
 - They get pinned on precipitates;
 - They get pinned by other dislocations or at dislocation network/tangles;
 - They effectively get pinned by splitting into partial dislocations (FCC)
- But pinned dislocations can also get released by,
 - Dislocation climb (edge dislocations);
 - Dislocation cross-slip (screw dislocations)
- In addition, dislocations get eliminated when they run into grain boundaries, but,
- Dislocations can also get created at steps on grain boundaries under stress.
- Dislocations can also arise via aggregation and collapse of vacancies (Fig.16).
- Dislocation density increases as strain accumulates. The main reason is believed to be that the passage of dislocations itself creates new dislocations at certain positions – essentially by interaction with other lattice defects. An example is a Frank-Read source, where a dislocation attempts to glide between two points which pin its ends – see Figs.17,18.

Figure 16: *Aggregation of vacancies creating dislocations*

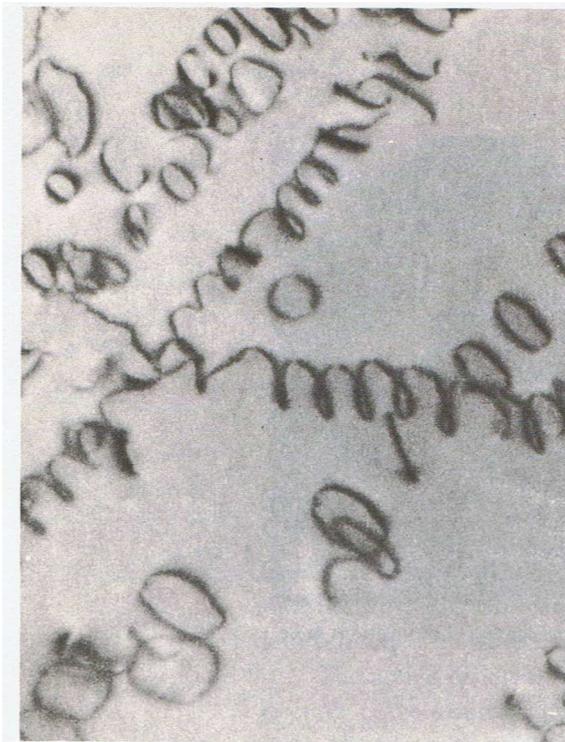


Figure 17 Electron micrograph of dislocation loops formed by aggregation and collapse of vacancies in Al-5% Mg quenched from 550° C. The helical dislocations are formed by the “climb” of screw dislocations as a result of the precipitation of vacancies. Mag. $\times 43,000$. (Courtesy of A. Eikum and G. Thomas.)

Figure 17: *Diagram illustrating how a Frank-Read source creates new dislocations*

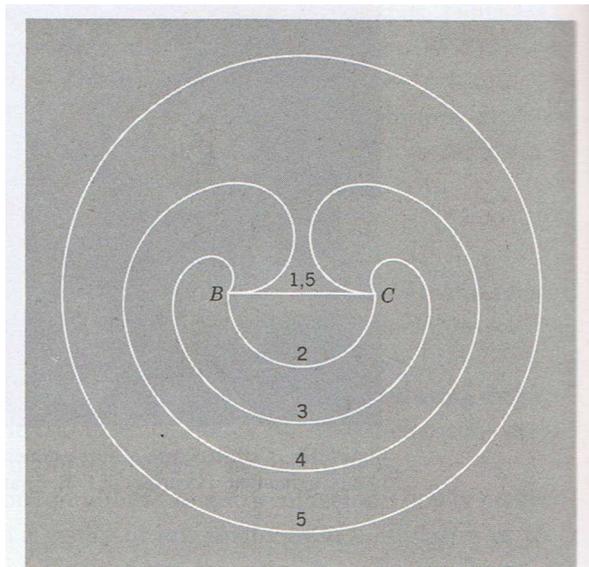


Figure 18 Frank-Read mechanism for multiplication of dislocations, showing successive stages in the generation of a dislocation loop by the segment *BC* of a dislocation line. The process can be repeated indefinitely.

Figure 18: *Micrograph of actual Frank-Read source (in Si)*

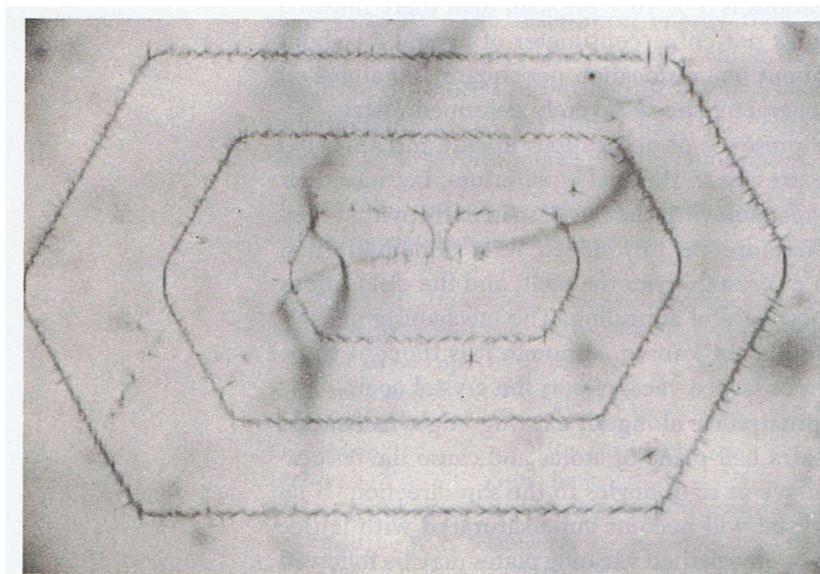


Figure 19 A Frank-Read dislocation source in silicon, decorated with copper precipitates and viewed with infrared illumination. Two complete dislocation loops are visible, and the third, innermost loop is near completion. (After W. C. Dash, *Dislocations and Mechanical Properties of Crystals*, Fisher et al. (eds.), Wiley, 1957, pp. 57-68.)

Qu.: So dislocation glide produces plastic and/or creep strains?

In effect, yes. But to be pedantic, plastic or creep strains should not be equated with pure glide of a dislocation, because glide is a conservative process. The glide could be reversed without energy loss. The irreversible characteristic of plasticity and creep is generated when a dislocation is created, annihilated or becomes pinned, or otherwise interacts with other lattice defects in such a way as to dissipate energy. Hence, for example, jog formation is irreversible (non-conservative). A jog is formed by the interaction of a dislocation and a vacancy, or the interaction of two dislocations.

Qu.: How do two dislocations interact when they cross?

Plasticity (e.g., cold work) and creep move dislocations around. Consequently this straining will cause dislocations to collide. Jogs or kinks are formed as a result of two dislocations crossing each other, as follows,

- Two orthogonal edge dislocations with perpendicular Burgers vectors interact to produce a jog in one of the dislocations;
- Two orthogonal edge dislocations with parallel Burgers vectors interact to produce two kinked dislocations;
- Two screw dislocations interact to produce jogs in both;
- An edge and a screw dislocation interact to produce a jog in the edge dislocation and a kink in the screw dislocation.

Qu.: How do jogs and kinks impede the motion of dislocations?

A kink in an edge dislocation is unstable since it is easily annihilated by glide.

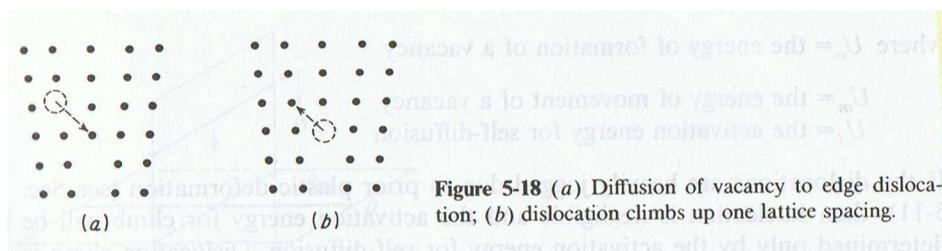
Neither kinks nor jogs in an edge dislocation impede its ability to glide. A jog in an edge dislocation will tend to persist whilst it glides, but without impeding the motion.

But the glide of a screw dislocation is impeded by jogs. The jog can only move freely (conservatively) in the direction parallel to the dislocation line. So glide of the dislocation can only occur by taking the jog with it, requiring climb/cross-slip. This is non-conservative and thermally activated. The result is that screw dislocations tend to move more slowly than edge dislocations.

Qu.: How do pinned dislocations get free? (Climb & Cross-Slip)

Edge dislocations can get around an obstacle by ‘climb’. This is the process by which the dislocation line can move up or down. The diffusion of a vacancy to the dislocation causes it to move up (positive climb), see Fig.18. Similarly, the diffusion of an atom to the dislocation (or the diffusion of a vacancy away from the dislocation) cause the dislocation to move down (negative climb). More details below.

Figure 19: Illustrating the climb of an edge dislocation



Screw dislocations can get around obstacles by cross-slip. This is simply their ability to glide on any plane orthogonal to the dislocation line.

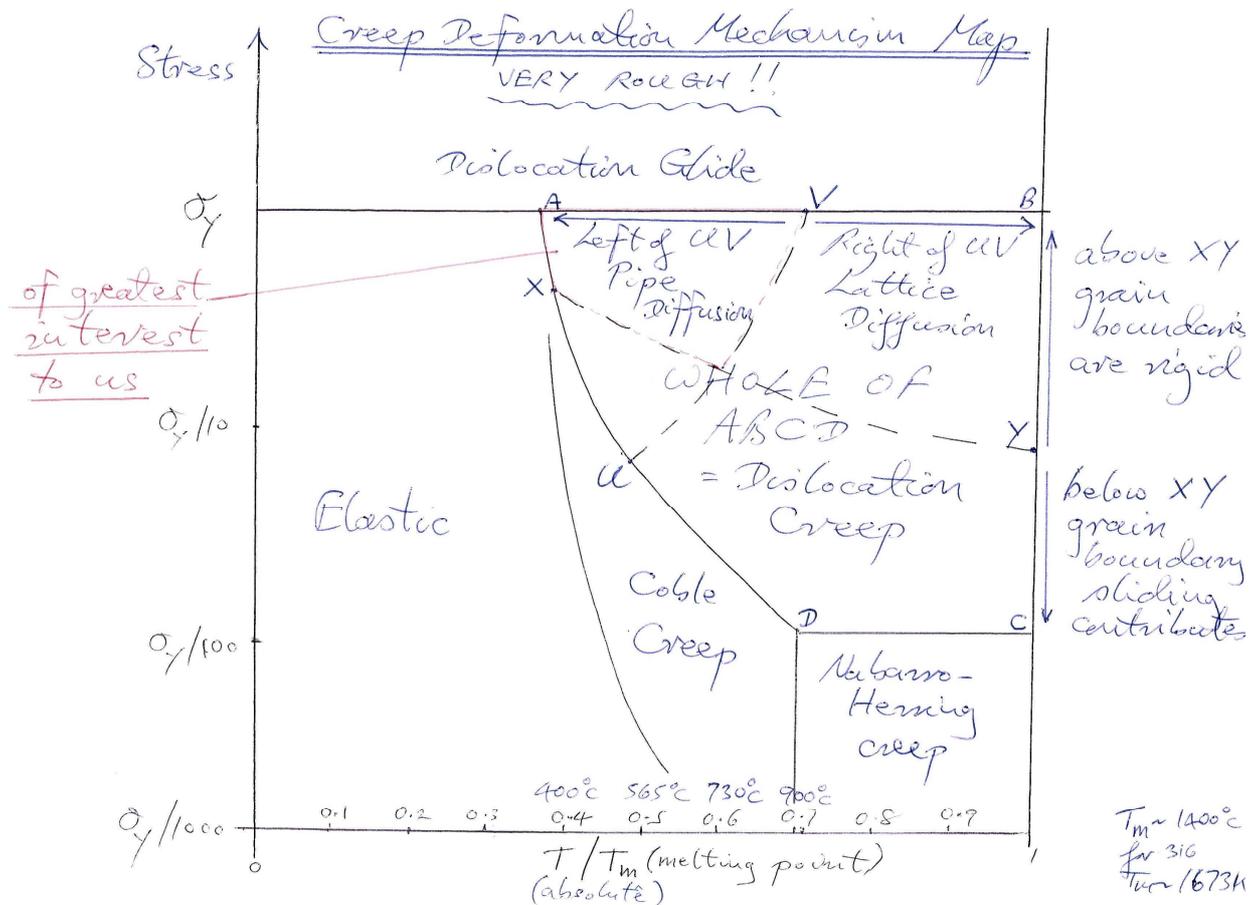
Qu.: So what are the mechanisms leading to creep deformation?

The standard nomenclature for creep deformation mechanisms are,

- Dislocation Glide: Slip of free dislocations. This determines the creep strain rate only at high stresses, in excess of yield.
- Dislocation Creep / Climb: Despite the name “dislocation creep”, the rate controlling process in steady state creep is often vacancy diffusion. This is because the creep strain rate can be the result of a balance between dislocations becoming pinned (hardening) and becoming free by edge dislocation climb. The principal rate-determining mechanism in dislocation climb is vacancy diffusion. However, the strain itself is still due to dislocation glide. Screw dislocation cross-slip can also contribute to dislocation recovery. Grain boundary sliding can also contribute to strain.
- Dislocation Creep / Solute Drag: In some cases the rate controlling process in dislocation creep is not climb but the rate of dislocation movement between pinned positions. This can happen if the viscous drag exerted on a gliding dislocation by the solute atmosphere is sufficiently great. There is some evidence that solute drag becomes the dominant determinant of creep rate in 316 at temperatures below $\sim 580^{\circ}\text{C}$.
- Diffusion Creep: Grain boundaries subject to tensile normal stress will attract atoms by diffusion, and compressive boundaries tend to loose atoms. Where diffusion occurs predominantly through the crystal matrix this is called Nabarro-Herring creep. This tends to happen only at very high temperatures. At lower temperatures, ‘pipe’ diffusion down dislocation network lines or along grain boundaries tends to dominate, known as Coble creep.

Qu.: In summary, what mechanisms control creep rates under conditions of most interest to us?

The stress/temperature conditions under which the above processes dominate creep deformation are illustrated in the mechanism map of Fig.20 for a generic material. We are most interested in dislocation creep in which deformation is dominated by dislocation slip and the rate is controlled by vacancy diffusion, predominantly pipe diffusion down dislocation networks and grain boundaries. Creep strain rate is controlled either by the rate of climb, which in turn is controlled by the rate of vacancy diffusion, or possibly by the dislocation glide rate throttled by solute drag.

Figure 20: Creep deformation mechanism map

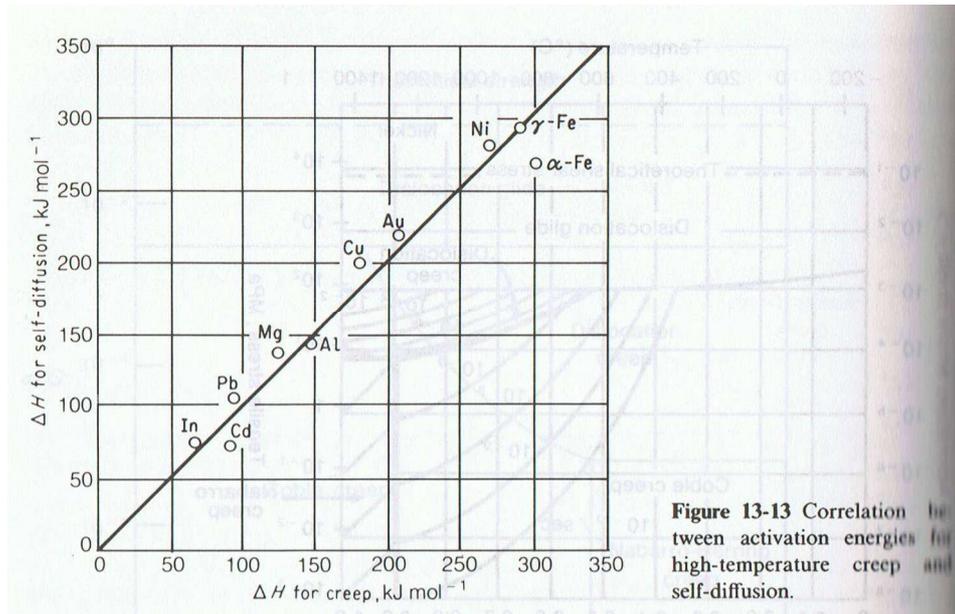
Qu.: What is the implication of climb control for creep rates?

Dislocation climb applies only to edge dislocations (since screw dislocations do not have an extra plane of atoms). The following summary of climb therefore applies to edge dislocations only...

- Positive climb: diffusion of vacancy to dislocation line – made energetically more favourable by imposing a compressive stress in the slip direction
- Negative climb: diffusion of atom to dislocation line (i.e, diffusion of vacancy *away from* the dislocation) – made energetically more favourable by imposing a tensile stress in the slip direction
- Both positive and negative climb can result in a dislocation escaping from an obstacle.
- BUT it is clearly not likely that climb will occur at all positions along a dislocation line all at once. If climb has occurred along one portion of a dislocation, up to point x, and no further, then there will be a step in the dislocation line. This is a **jog** at position x.
- Climb occurs preferentially at an existing jog.
- The activation energy for dislocation climb is thus $U_{\text{climb}} = U_j + U_d$, where U_j is the energy for jog formation and U_d is the activation energy for self-diffusion. But if the jogs are pre-existing, then the activation energy for climb is the same as that for self-diffusion, $U_{\text{climb}} \approx U_d$.

- If climb is the rate controlling process for creep, then the activation energy for creep is expected to be the same as that for climb, which in turn is the same as that for self diffusion, i.e., $U_{\text{creep}} \approx U_{\text{climb}} \approx U_d$. This does seem to be born out in practice (at least for some materials under some conditions).

Figure 21: Illustrating that creep is controlled by self-diffusion for a wide range of materials. The activation energy for creep deformation equals the activation energy for diffusion to a good approximation for many materials. Since climb is diffusion controlled, this is consistent with creep being climb controlled.



Qu.: How does the primary-secondary creep transition occur?

One way of explaining the faster initial creep strain rate (primary creep) is that there is initially a certain density of mobile dislocations which are free to glide. As these move and become pinned, the number of such mobile dislocations reduces. This is the mechanism of hardening (reducing strain rate).

The steady secondary strain rate is achieved when the rate at which pinned dislocations are being released by climb or cross-slip equals the rate at which mobile dislocations are being pinned by gliding into obstacles.

Note that although the primary creep reduces the density of mobile dislocations, the total number of dislocations increases substantially due to the dislocation creation mechanisms discussed above. Consequently, the number of dislocations climbing/slipping around obstacles increases as primary creep proceeds simply because the number of pinned dislocations increases.

The above description continues to apply (I think) even if the creep rate is controlled by solute drag. “Solute drag” is the regime when the viscous drag of the solute atoms on the gliding dislocations is the rate determining step. This means that the rate at which dislocations get pinned is reduced. Hence, the rate at which pinned dislocations are required to climb around obstacles to produce a steady secondary creep rate is also reduced. This means that lower temperatures are required. Hence, solute drag applies at the lower temperature end of “dislocation creep”.

Qu.: How does plastic pre-strain influence subsequent creep rates?

The effect of cold work, or other forms of pre-strain, on subsequent creep behaviour is often very marked. In 316, the initial (primary) creep rates tend to be reduced by pre-strain. But pre-strain can have a seriously adverse effect on creep ductility.

- The most readily apparent means by which plastic pre-strain might affect primary creep is by using up (some of) the initial supply of mobile dislocations. This would be expected to lead to a reduced primary creep rate.
- We have seen that secondary creep is controlled by climb, and that climb occurs preferentially at jogs. Hence, secondary creep rates will depend upon the jog density. How do jogs arise?
- The energy for jog formation is $U_j \sim 1\text{eV}$. If jogs arose thermally, due to random thermal motions, the fraction of the lattice positions along a dislocation line which are jogged would be roughly $\exp\{-U_j / kT\}$. This would lead to very few jogs being formed by purely thermal means at usual temperatures.
- In practice, most jogs are created by plastic/creep deformation. This is therefore another means by which cold work can influence subsequent creep rate. Note that both tensile and compressive pre-strain would therefore be predicted to influence subsequent creep (which is true). The creation of jogs by pre-strain would therefore facilitate climb and hence be expected to increase the secondary creep rate.
- But plastic strain also pins previously mobile dislocations – so the effect of pre-strain on creep rate can go either way - perhaps reducing primary rates but increasing secondary rates. Is this true? (Don't know).

Qu.: Why are austenitics more creep resistant than ferritics?

This is obviously an important question but the answer is less clear cut than I would like. I can offer three possible reasons. The first is the simplest, and frankly is the only one I really understand,

Self-Diffusion Rate

Dislocation creep relies largely upon vacancy diffusion to drive the climb of pinned dislocations. But the diffusion coefficient through alpha-iron (BCC, ferrite) is far greater than that through gamma-iron (FCC, austenite), by a factor of ~ 350 at the ferritic/austenitic transition temperature. Consequently, the creep strain rate is far greater for ferritic steels than austenitic steels at the same temperature and stress.

The reason for the faster diffusion rate in the BCC structure is that the activation energy is lower than for FCC (see Fig.21). I presume this is simply because BCC is less closely packed than FCC. The looser structure will naturally encourage faster diffusion rates. (Is it as simple as that? Not sure).

This is one reason why austenitic steels (FCC) have higher creep resistance than ferritics (BCC). However there is also a more complicated reason which is often cited...

Low Stacking Fault Energy in FCC Implies Reduced Creep Rate

Deep breath, here goes...

- Slip of the [111] planes in FCC can locally corrupt the ABCABCABC stacking sequence, locally resulting in very thin ABABAB regions. The energy required to do this is the “stacking fault energy”.
- Austenitic stainless steels can have low stacking fault energy, but this is sensitive to composition.
- Alloying elements of high valence (Cr, Ni) further reduce stacking fault energy.
- Stacking faults in FCC are bounded by partial dislocations. (NB: a pair of partial dislocations have burgers vectors which add vectorially to be equivalent to a full dislocation).
- The lower the stacking fault energy, the broader is the stacking fault ribbon.
- Normally, screw dislocations in FCC do not have a preferred slip plane. For this reason they can usually ‘cross-slip’ to get around obstacles.
- But when decomposed into a pair of partial dislocations (i.e., a stacking fault) a screw dislocation *is* then constrained to a particular slip plane. Consequently...
- Partial screw dislocations cannot cross-slip to avoid obstacles, and this leads to a reduced propensity to creep. This is the mechanism whereby low stacking fault energy leads to reduced creep strain rate (and, presumably, to increased creep strength).

No, don't ask me any questions on that. Consult the texts.

There is also one further mechanism which leads to increased creep resistance of austenitics...

Lomer-Cottrell barriers

In FCC, slip on two intersecting [111] planes (duplex slip) results in a barrier which makes the dislocations immobile (they combine to form a dislocation which is not on the easy-slip plane). This also leads to a reduced creep strain rate (hardening).

Qu.: What is creep damage?

We do not know what creep damage is in metallurgical terms.

For ferritic materials, creep damage can usually be assumed to be cavitation.

For austenitic materials, creep damage is also sometimes cavitation. However there appear to be cases when damage leading to local material failure occurs but without cavitation being apparent (e.g., in lab crack growth specimens). However even this is uncertain because it is not easy to identify cavities. They are easily confused with non-metallic inclusions (e.g., carbide precipitates). Whether a sample is declared as cavitated or not may depend on which metallurgist you get to look at it!

However, we shall discuss the formation and growth of cavities since this is certainly one form of creep damage. Cavities can form both within grains and on grain boundaries. Cavitation on grain boundaries is generally most significant under the conditions of interest to us, both in austenitic steels and low alloy ferritic steels. (NB: Type IV cracking in inter-critically refined HAZ, as well as coarse HAZ cracks, arise

from grain boundary cavities – and it is grain boundary cavitation which is quantified when replicas of site welds are examined).

Austenitics

In austenitics, the process of damage starts with the formation of sites on the grain boundaries at which cavities can nucleate. These sites are energetically favourable for vacancies. Hence, diffusion leads to the accumulation of vacancies at these nucleation sites and a very small cavity forms. The imposition of stress assists in driving more diffusing vacancies towards the micro-cavities, causing them to grow. When the size of the cavities approaches their distance apart, they interlink and cause the whole of the grain boundary to decohere. This is a micro-crack of grain size (perhaps 50-200 microns in typical 300-series structures/weldments). The linking together of several such micro-cracks forms a macro-crack. It is frequently found that creep cracks in structures have disconnected micro-cracks ahead of the main crack tip, and also some grains with heavy cavitation which have not yet linked up – essentially catching this process in action.

A later session will develop algebraic models based on these ideas. One of the implications of this mechanism is that creep ductility depends mostly on the spacing of the nucleation sites on the grain boundaries. These models also suggest a mechanism underlying the stress-state dependency of creep ductility.

Ferritic Weldments

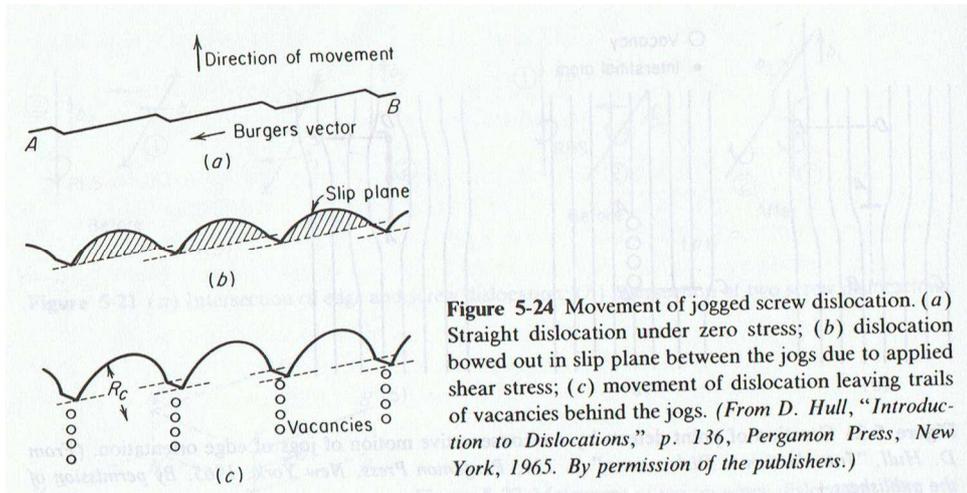
The most significant feature of creep damage in low alloy ferritic weldments (e.g., CMV, 2.25Cr1Mo) is that it is highly inhomogeneous. Damage again takes the form of cavitation in most cases, but generally accumulates in certain weldment zones. The HAZ is generally the most susceptible. Coarse HAZ is perhaps the most susceptible to cavitation damage. This should be reflected in assessments of damage, if they are to be indicative. However, due to alternating bands of coarse and refined HAZ, many such weldments act as if they had intermediate (mixed) HAZ properties. In this case, the most susceptible region might be the intercritical region (details in a later session). Cracks forming within the intercritical region are called Type IV cracks. These are sufficiently important to us that we have techniques based on replication for counting Type IV cavity densities and correlating these with life usage (discussed further in a later session).

Qu.: How do vacancies get created?

We have seen how important vacancies are to creep processes. We have also seen that vacancies have a sufficiently modest activation energy to be formed to some extent from available thermal energy. But this equilibrium density of vacancies is insufficient on its own to result in creep failure. So vacancies must also be created by other mechanisms.

- [1] We have seen that negative climb of edge dislocations will create vacancies (see Fig.19).
- [2] Vacancies are also produced when screw dislocations move through each other. Jogs will be formed along the length of a given screw dislocation where it crosses another screw dislocation. Its motion is impeded at those points, leading to a serrated shape, see Fig.22. If there is sufficient force to continue to move the dislocation, it will tend to leave a trail of created vacancies behind it as the jogs are forced to climb. See Fig.22.

Figure 22: *How vacancies are produced by intersecting screw dislocations*



Qu.: What is the mechanism causing tertiary creep?

Tertiary creep might be attributed to the reduction of load carrying cross-section as deformation proceeds. This could be simply a true-stress effect. On the other hand it might be due to loss of effective area on the grain-scale. The build-up of cavities on grain boundaries will lead to reduction of load-carrying area and hence increasing local true stress at constant load. Once started, this leads to a rapid positive-feedback in which increased true stress causes increased strain rate, which causes increased cavity growth rate, which causes increased rate of increase of stress, etc. It is easy to see that failure will ensue quickly once this vicious circle kicks in.

However, this is not the only possible mechanism for tertiary creep. An awkward fact is that metallurgical examination often reveals very few cavities even in material which is unambiguously in the tertiary regime. So, an alternative explanation is that tertiary behaviour results when the matrix material becomes thermally softened (thermal ageing). The mechanism is not clear (but see below).

Qu.: What is thermal softening?

Ex-service 316H, if subject only to modest service stressing but temperatures above 500°C, becomes softened. It shows faster creep rates than virgin material when removed from service and tested under higher stress than it saw in service. The mechanism is not unambiguous but may be as follows. The high temperature can cause pinned dislocations to be released by climb. If the stress is modest, not all these freed dislocations will be used up in causing creep strain (i.e., they do not all subsequently become pinned again). Hence, the density of mobile dislocations can increase over time. This is essentially the reverse of the hardening that normally occurs in primary creep, in which the number of mobile dislocations is reduced. So this proposed mechanism only makes sense if the stress is low and the amount of creep in service is small.

Hence, thermal ageing can result in creep-softening. But the above proposed mechanism cannot account for tertiary creep since this occurs after substantial creep straining. A mechanism which could lead to softening after substantial strain accumulation is the increasing density of pinned dislocations, jogs and vacancies which result from the strain. All these will lead to increasing strain rates due to increasing dislocation recovery due to climb.

Qu.: What does all that amount to?

Creep is a very complex mixture of interacting mechanisms

You cannot predict the creep behaviour of a material unless you know its history

Much of what is often attributed to “scatter” in creep data is probably a real reproducible effect, if sufficient were known about the history of the material sample

Many of the mechanisms discussed above can be turned into algebraic models of the underlying physics. But these are usually so approximate that they are not predictive in any useful sense. Nevertheless they can be helpful in explaining qualitative features, or in suggesting suitable algebraic forms for fitting experimental data.

Qu.: Is ductility exhaustion the right approach to assessments?

Consider the following excessively naïve view,

- 1) Creep deformation arises due to dislocation glide, but,
- 2) Creep failure is due to vacancy diffusion causing cavity growth.

This could lead you to conclude that deformation and failure were unrelated. In which case, why should failure occur at a ‘critical’ deformation, the ductility?

The purpose of this session has been to show that things are not so simple. The movement of dislocations can require vacancy diffusion, and the movement of dislocations can generate new vacancies. So the mechanisms (1) & (2) are not independent.

Our interest in creep deformation is largely because it is assumed that there *is* some sort of correlation between deformation and damage leading ultimately to failure. We are rarely interested in deformation *per se*. (Although creep strain rate does feature in an important way in creep crack growth assessments, this is essentially just because creep strain is assumed to be related to creep damage near the crack tip). However, the complexity of the interacting creep mechanisms is such that the nature of the correlation between deformation and damage might be very complicated. It does seem to be...

Although creep deformation mechanisms and creep damage/failure mechanisms are intimately intertwined, there is no reason to expect failure to occur at a fixed value of creep strain.

In other words, “creep ductility” is not a material property but will vary according to the sequence of conditions to which the material has been exposed.

Any assessment procedure based upon ductility exhaustion must therefore include a means of taking account of the effect of the history and service conditions on “creep ductility”, e.g., the stress state dependence. This is the current R5 approach.

An alternative would be to drop the concept of creep ductility and formulate damage in terms of some alternative "state variable" to strain (e.g., stress or energy density or dislocation density). But dependence of damage on the historical sequence of conditions is unavoidable whatever formulation is adopted.

An advantage of the ductility exhaustion approach is that it focuses attention on low ductility failure modes, e.g. reheat cracking and creep crack growth. Whilst such failure modes could possibly be predicted via a stress-based approach, an excessive reliance on stress-rupture without adequate account of multi-axiality and HAZ behaviour can lead to debacles like the P91 saga and our own austenitic reheat cracking problems.