

Lead-Free Solder Joint Reliability

Oxymoron?

(I like it, but ...)

Peter Borgesen, AREA Consortium

borgesen@uic.com

607-768-2132

pborgese@binghamton.edu

You've got a tough job, and I'm here to make it harder

If you pay close attention, you'll learn how to make your tests predict an order of magnitude higher/lower life or *favor the wrong solution* without anyone noticing.

'Tools' include:

- Assume IPC/Engelmaier expression or Norris-Landzberg based models for thermal cycling life

- Assume Weibull (espec. 3-parameter) for thermal or isothermal life

- Vary pad surface roughness?

- Vary reflow profile (interactions with alloy)

- Select time between reflow and test

- Random vibration test?

- Vary low-T dwell

- Vary high-T dwell

- Combine tests, vary order!

- Assume Miner's rule (worse than you think)

- Vary solder volume

.....

I can't be responsible for what you do with the knowledge (guns don't kill people, ...)

More to the point: Your test protocols are not only wrong, they are incomplete.



'Predictions are difficult, particularly about the future'

**R. Storm Petersen
Danish Philosopher**

What level of reliability do you care about:

- Life in service?

- Good enough?

- As good as last time?

- Best of alternative designs, materials, processes, ...?

In either case knowing that what we do is wrong, or at least imperfect, is not useful by itself:

- What is the best we can do right now, and how?

- How confident can we be about the conclusion? Would you bet \$100, \$1M, your life, ...?

- And for the longer term, what (research) plan will help alleviate this?

This presentation will not address whiskers. In terms of the other no-Pb reliability concerns we are better off, but to a limited degree:

- I'd bet \$100 on your approach, \$1,000 on mine, but ***not*** anybody's life -- **be afraid, be very afraid!**

If you aren't making cell phones, lap tops, game consoles, ... you are pretty much on your own (not a lot of real friends).

Even if you can wait for it, current level of industry funding & research may 'never' get us to where I'd bet anybody's life on lead free assessment.

There simply is not enough of a competitive advantage (\$\$) to solving it, but there is a societal interest in safety of certain products.

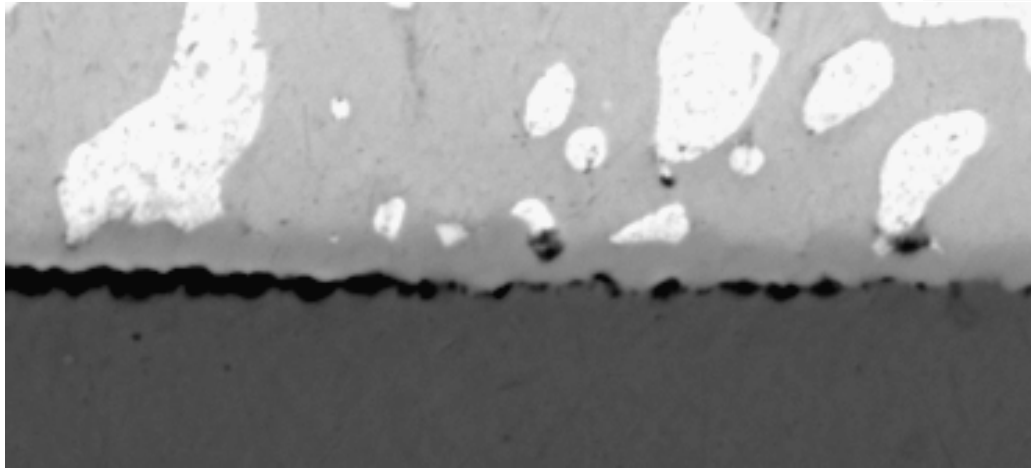
We need to rely heavily on in-depth mechanistic understanding established at the university (and government funding).

We need to know how to work with the university to get that, and translate it into practical consequences and guidelines.

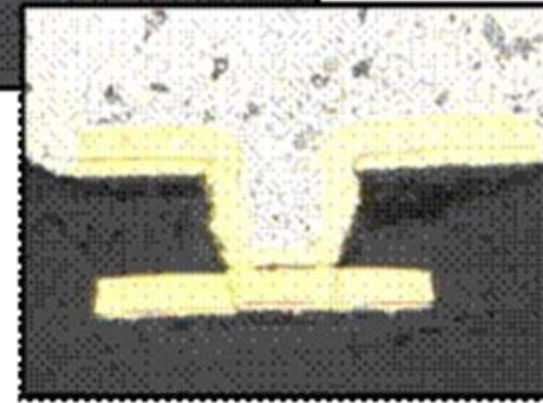
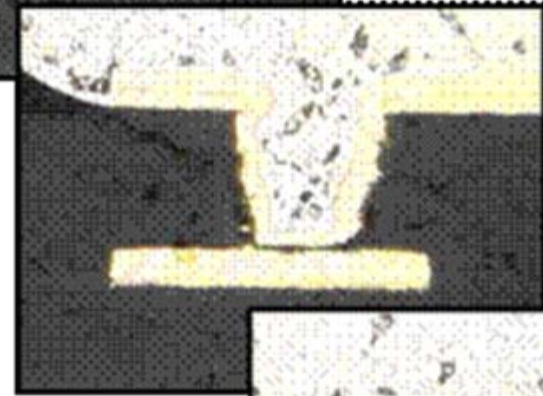
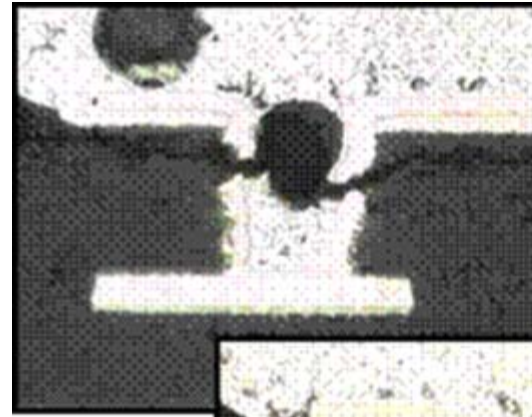
Lead Free Solder

Harder (less ductile):
enhanced cratering or IMC failure
greater sensitivity to defects

Higher temperatures:
Thicker IMC, more crater damage



"New" IMCs?



But this is not where the real challenges lie:

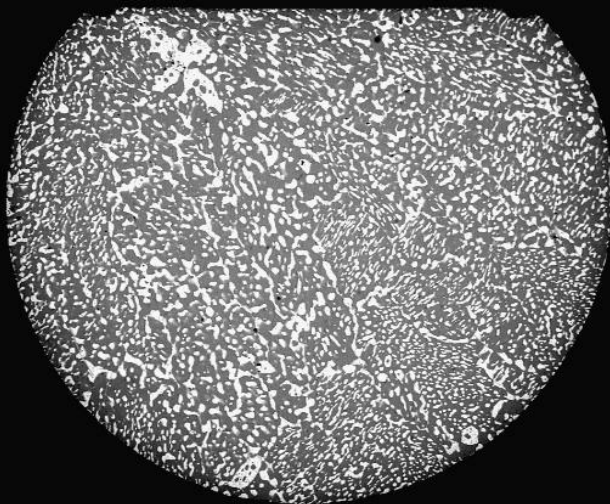
Lead free is so fundamentally different from SnPb that lots of experiences, rules, protocols, ... often become misleading.

Most of the time it may just look like a high temperature variant, but things that didn't matter before now do and *surprises* abound.

- Materials properties change with **solder joint size**, pad finish, process, history, alloy ... It has become much more difficult to generalize ('scale') results and experience.

Eutectic SnPb solder: composite of Sn- and Pb- regions

SnAgCu solder: few large Sn grains with small Ag-Sn & Cu-Sn precipitates



Acc.V Spot Magn Det WD | 50 µm
25.00 kV 6.0 350x BSE 11.8



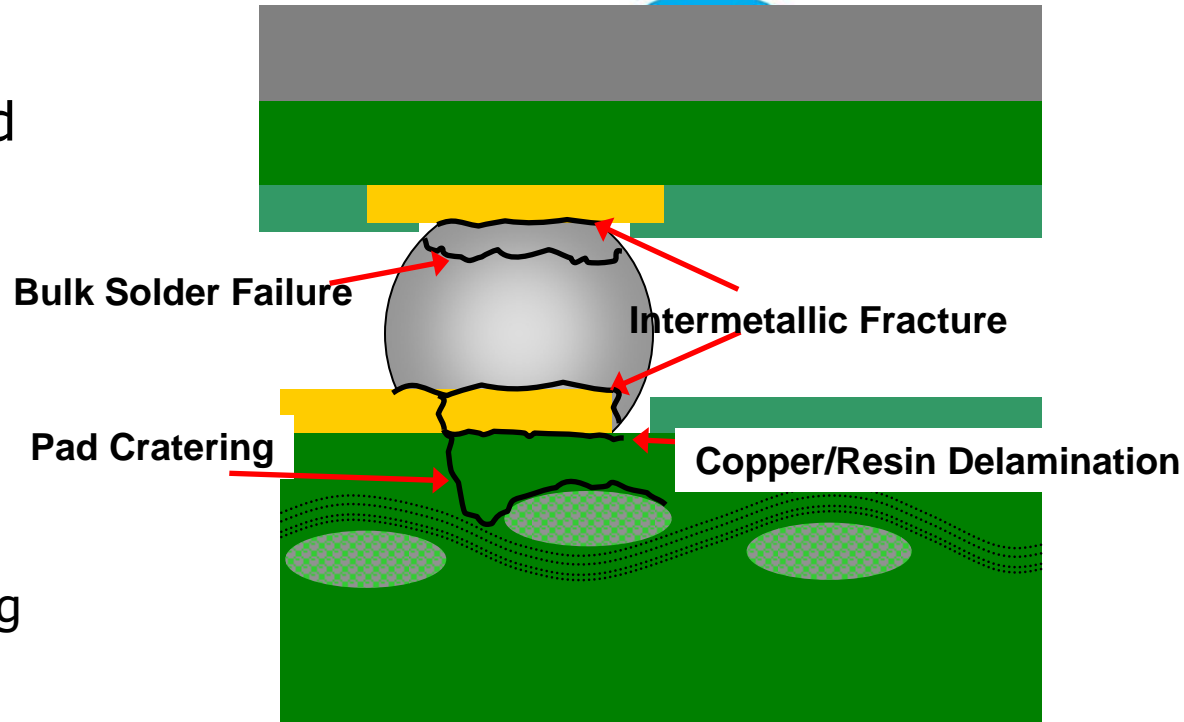
200 µm

The fact that test results can be organized (systematic trends discerned) makes us feel like we *understand* them, but ...

Extrapolation of accelerated test results to *long term service* was always risky, but by now somewhat 'calibrated' for SnPb. For lead free all we have is 'faith' (and what may be good enough for consumer electronics ...)

- Current thermal cycling tests are potentially *very misleading*. They may often even favor the *wrong alloy or process*. However, 'better' tests may often be impractical (require much too long dwell times, too little acceleration)?
- We have no way of accounting quantitatively for serious effects of long term aging on strength, vibration resistance, thermal fatigue life, ...
- Spectacular break down of Miner's rule is not only a problem for combinations of tests (but only noticed there).
- Inherent variability of solder microstructure provides for statistical 'outliers' (weak joints) commonly missed in current testing.

Not as easily generalized and extrapolated as we'd like for SnPb either



Even for a given type of loading
(e.g. repeated drops):

competing and interacting damage mechanisms and failure locations

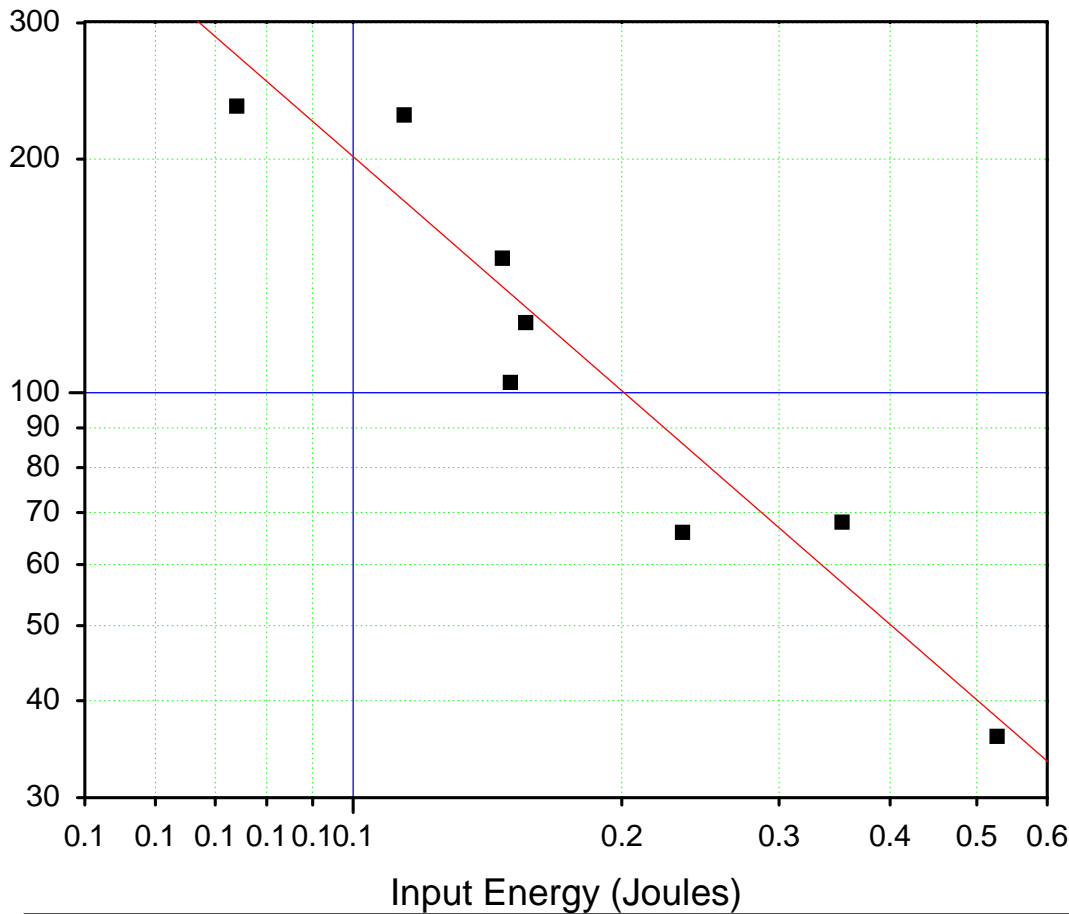
empirical test protocols may easily be strongly misleading as far as actual service conditions are concerned, or they become too large

-- repeated drops from different heights or at different temperatures or after different preconditions (time, temp., humidity) may rank designs, materials, processes differently

Scaling?



Generalizing results requires us to find a physically credible way to scale. Cratering under no-Pb joints seems to scale with input energy in drop?



$$N_f(63.2) = 20 * E^{-1}$$

Cycles to fail relates to the input energy through a power law.

E = calculated energy transverse to board.

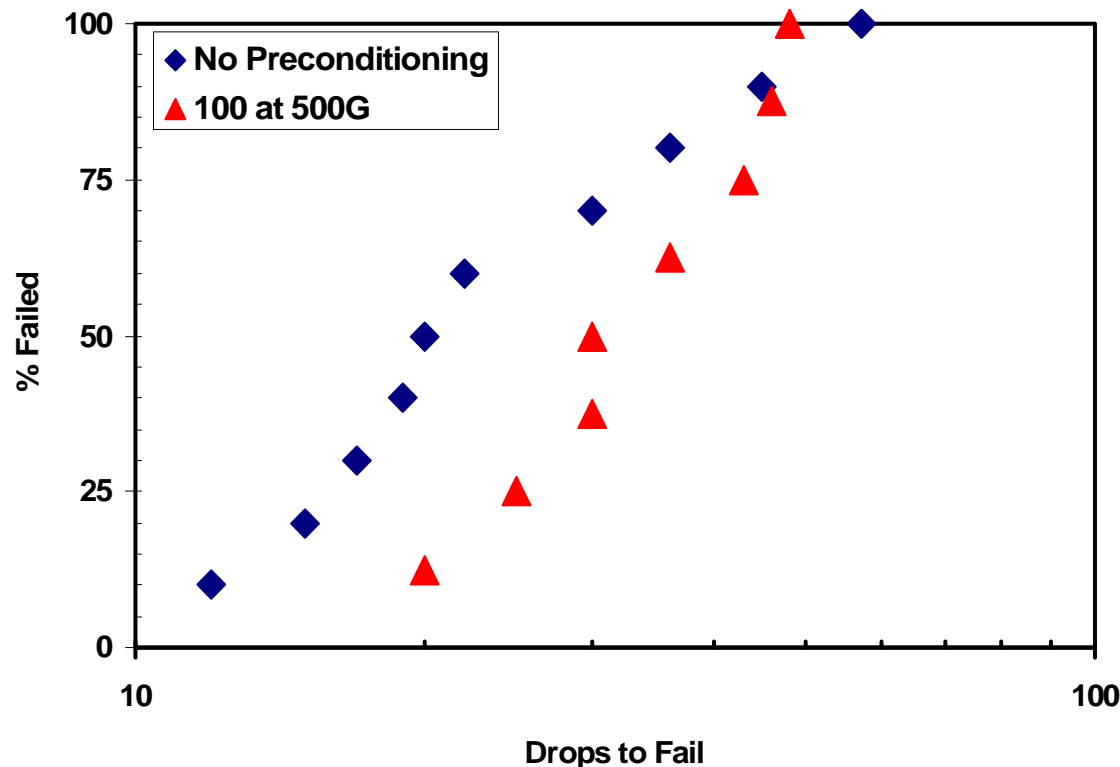
No trend of over/under estimating life.

However, this scaling *cannot* be general:

Scaling difficult without Palmgren-Miner linear damage accumulation hypothesis, but

- Drop: 1500-G input to fail by cratering
- Pre-conditioning: 100 drops at 500-G

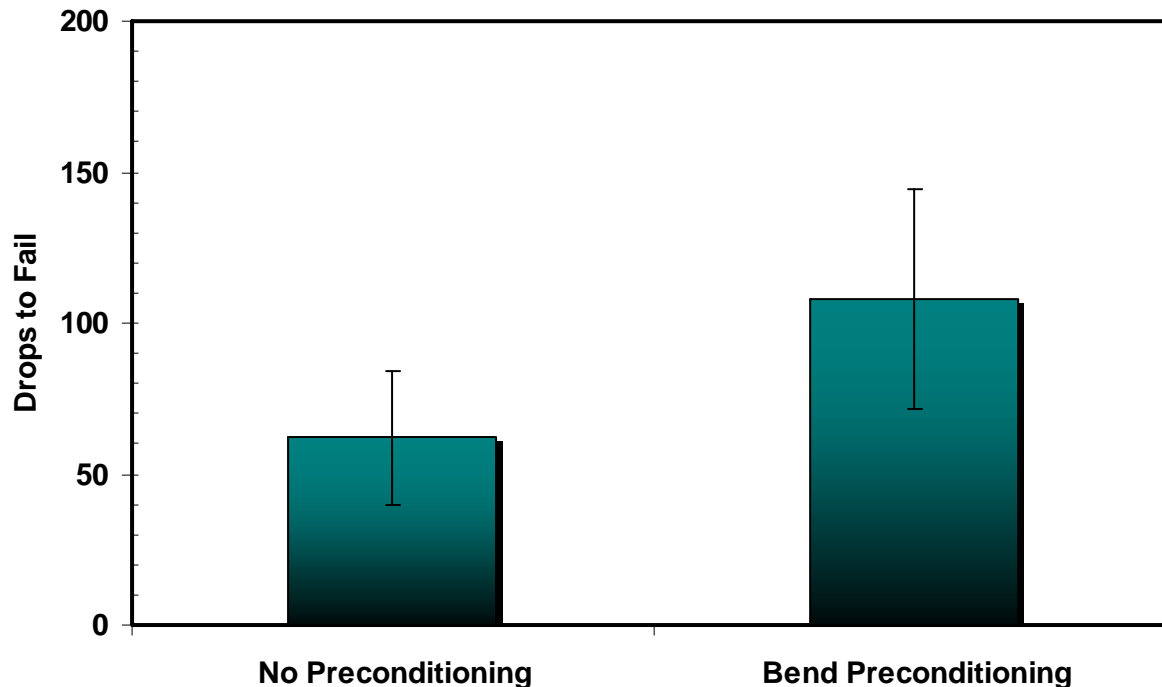
50% increase in drops to fail (N_{63}) due to pre-conditioning
(100 drops reduced hardness measurably)



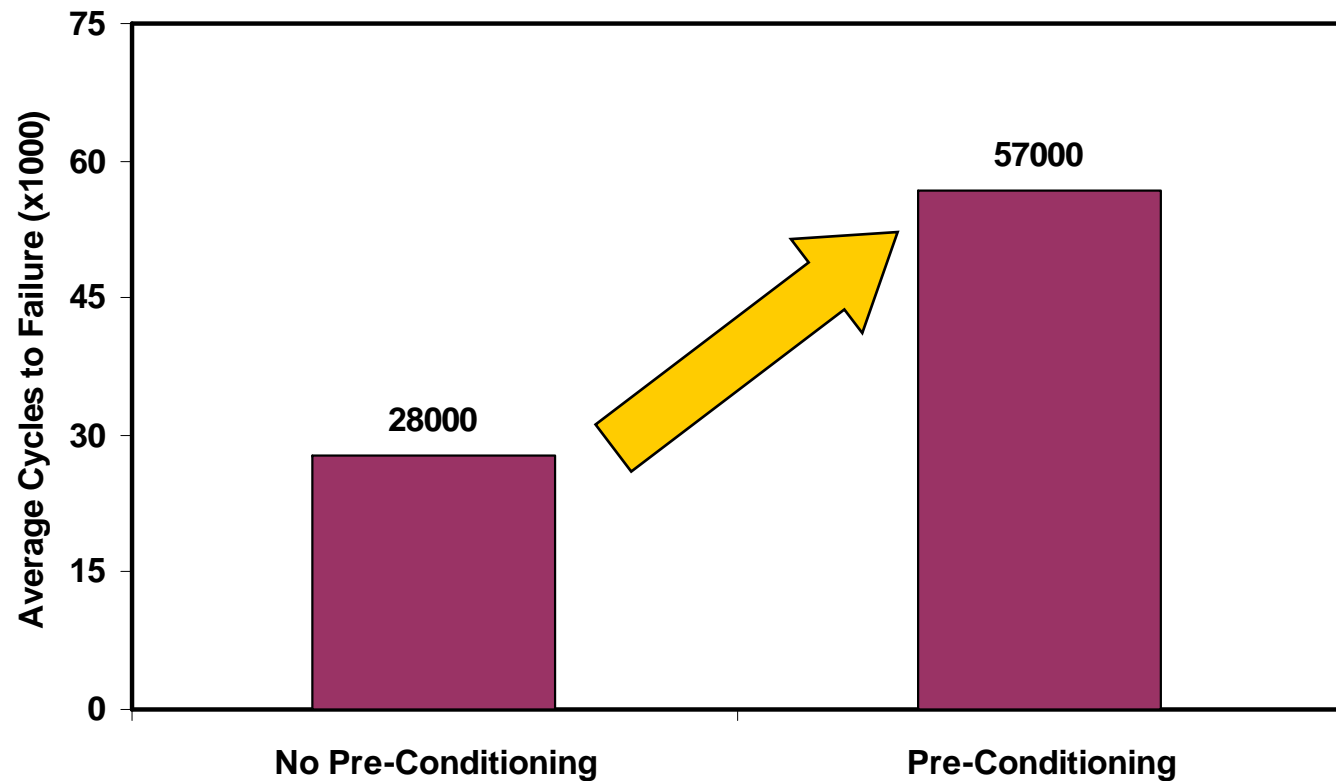
Drop: 500-G input to fail

Pre-conditioning: 100,000 cycles of low-amplitude Bend

Almost 2X increase in drops to IMC fail due to pre-conditioning (solder softening)



Something similar is observed for solder failure in vibration:
220,000 cycles of low level input (5-G) vibration gave 2x solder fatigue life in subsequent 20-G input test.



Does random vibration testing underestimate damage?

The complete failure of Miner's rule is clearly associated with dynamic softening of the solder:

In a *load* controlled shear fatigue experiment the result was opposite – preconditioning allowed for greater displacement and thus faster failure than predicted.

The enhanced life under combined loading may sound like good news (even potential for ESS applications?), but the reverse is also expected – shorter life than predicted under other combinations.

More generally, even in a single cycling test solder properties are different in each cycle. **Should be accounted for in modeling, but we need the input first.**

We propose 'backing out' state variable and crack length evolution laws from series of combined tests.

Life in service?

We weren't actually ever on a very sound mechanistic footing for SnPb. Only 'calibration' to decades of experience (fuzz factors) kept us from getting fired for our 'predictions': They became either reasonable or conservative for SnPb.

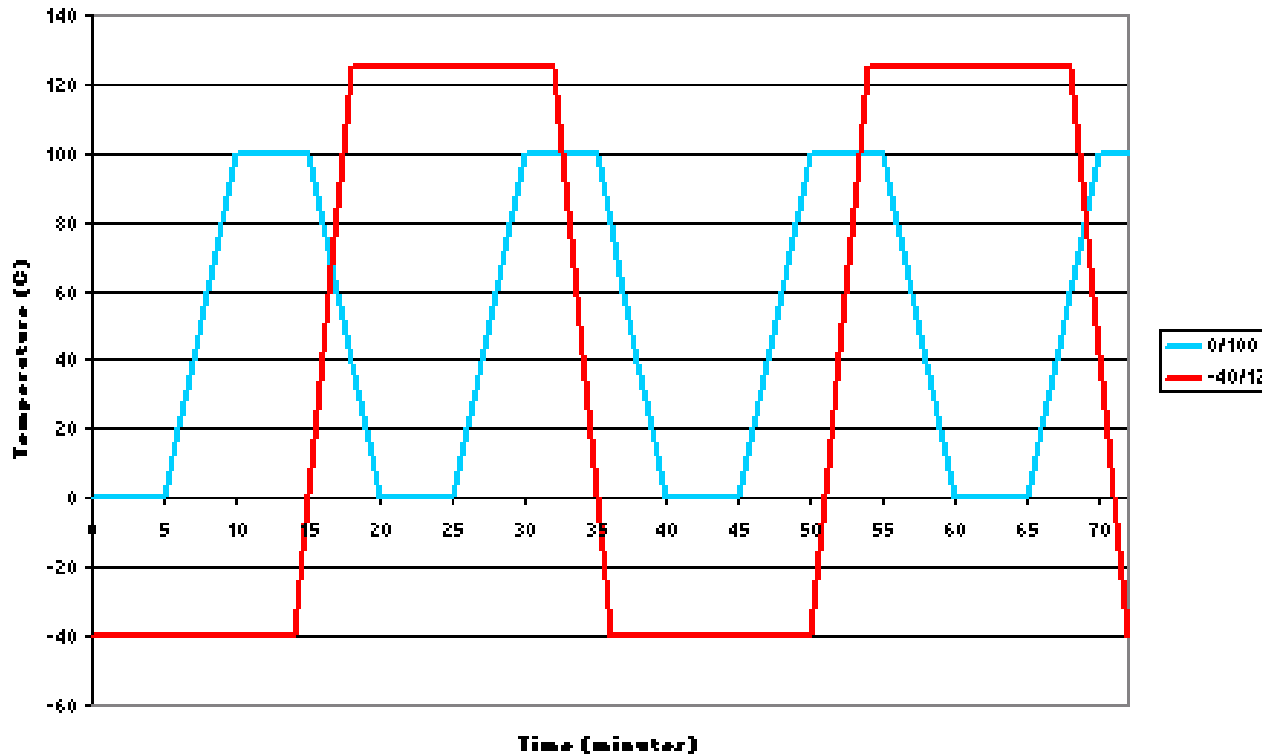
Anyway, we should ***certainly*** become concerned when

- i. we need greater accuracy (go closer to 'the edge')
- ii. loading conditions change fundamentally
 - semi-empirical scalings may not apply at higher/lower temperatures (extrapolations ...!)
 - underfilled systems are ***completely*** different, apparent modeling successes are deceptive
- iii. materials behavior changes fundamentally
 - **lead free solders!**

Thermal Cycling of no-Pb



Life in service: Need to know how to scale with *temperature*,
temperature *range*, *ramp rates*, and *dwell times*



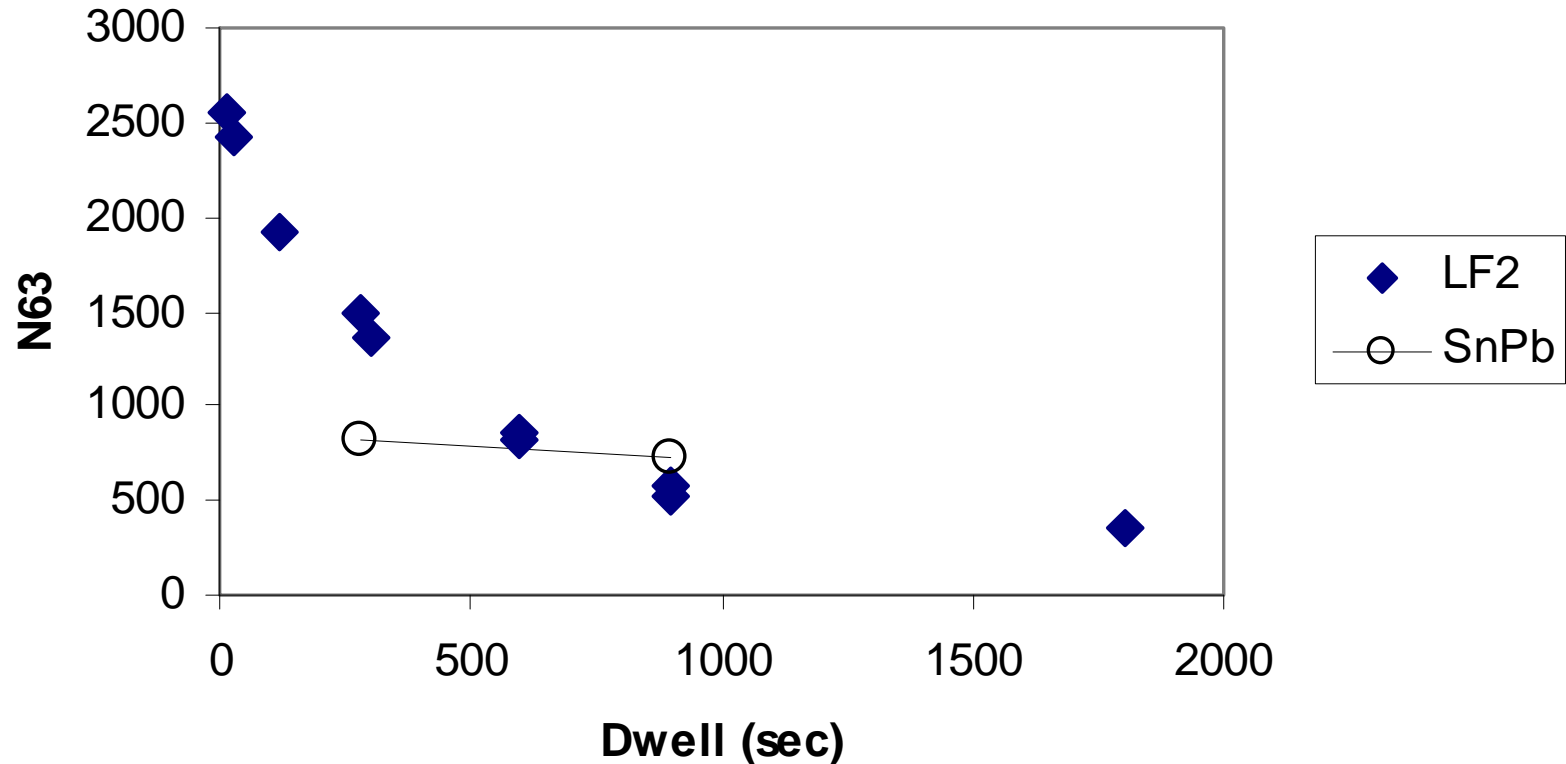
Comparisons: At least need same *ranking* as in service. Even that is clearly less trivial for lead free!

Effect of Dwell in Cycling

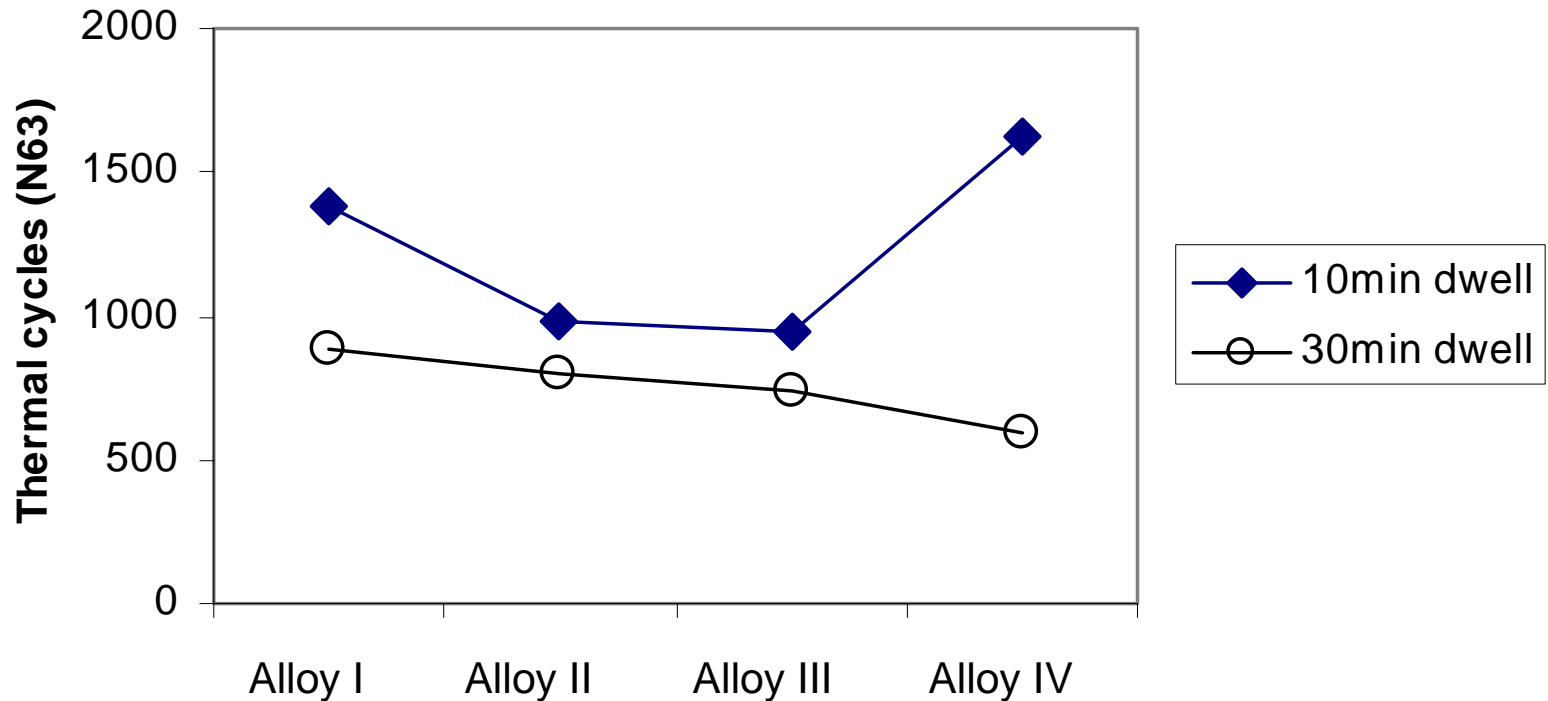


We are used to dwell time not being a critical factor (for SnPb) – just count cycles to failure and account for temperature range (and ramp rates)

0-100°C cycling



However, for lead free extending dwell from 10 minutes to 2 hours *can* reduce number of cycles to failure by 3-4x or more. **Magnitude of effect varies strongly with alloy, other cycling parameters, ...**



Comparison between lead free alternatives (designs, materials, processes)?

Short cycles may give wrong relative ranking: You may need very long tests to decide what will be best in service?

Several groups have reported dwell time effect (although usually without concluding that SnPb life is much greater than SAC). However, few have suggested what to do about it (test with short dwell, then correct):

Pan et al. (SMTAI 2005) propose a modified Norris-Landzberg equation with life proportional to an inverse power dependence, $N_F \sim t^{-b}$. Both they and Zhang & Clech (SMTAI 2005) report data that agree with $b = 0.136-0.156$.

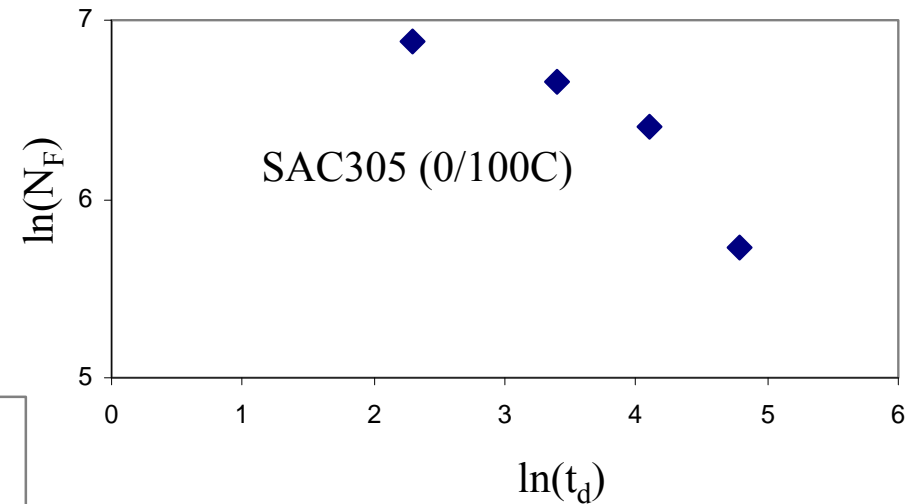
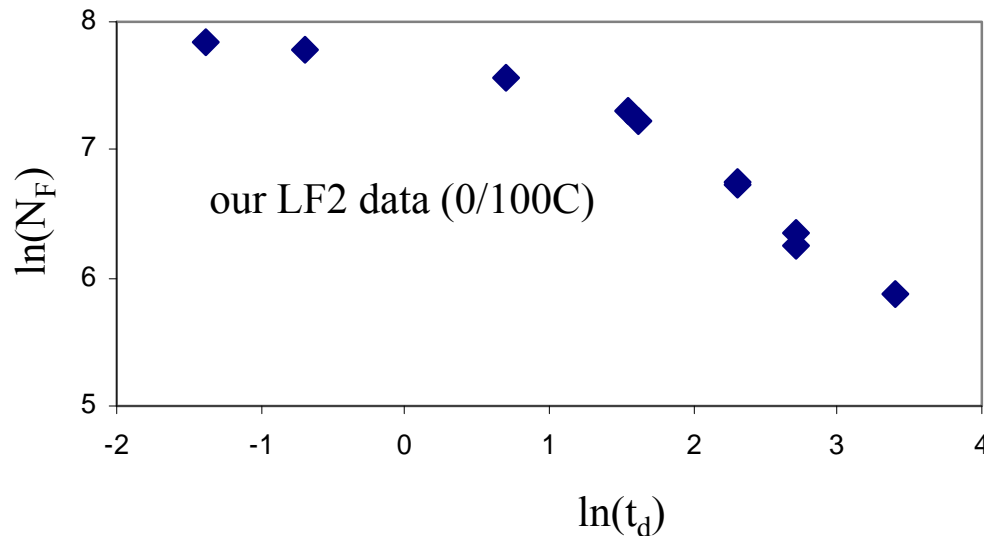
They only considered two t_d values, so fit to experimental data was not strong test of dwell time dependence. If we consider more ...

Effect of Dwell in Cycling



If power dependence is OK, we should get straight line on log-log plot:

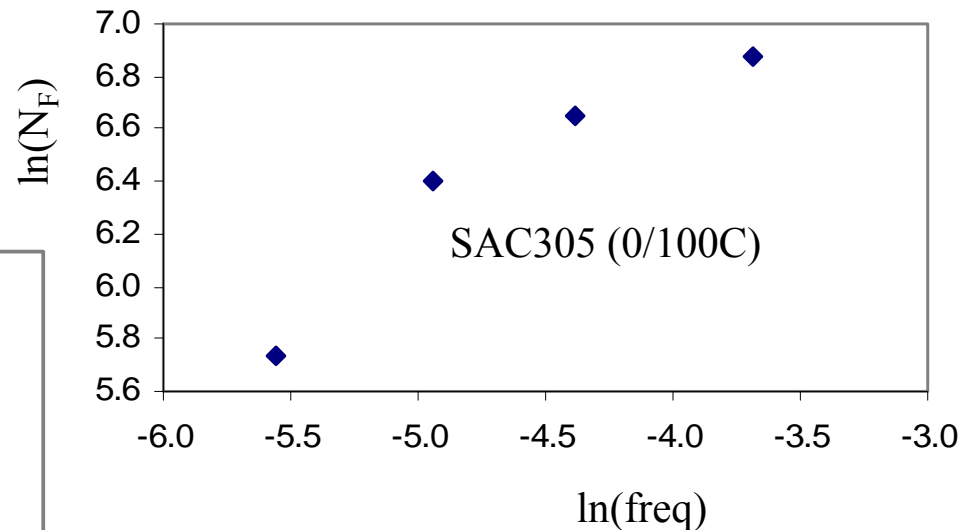
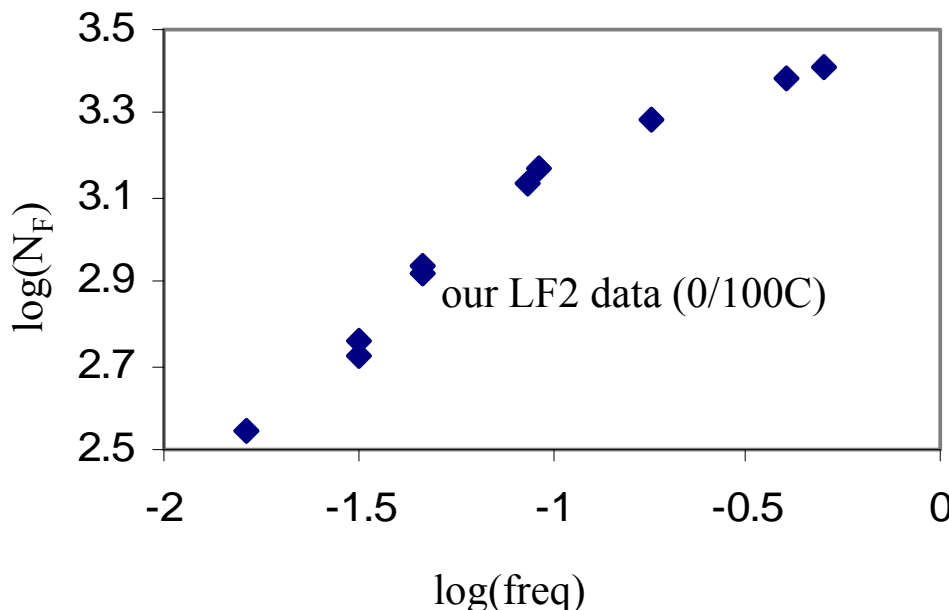
Our data do not follow power dependence



Effect of Dwell in Cycling



Vasudevan & Fan (ECTC 2008) propose a modified Norris-Landzberg equation with life proportional to an inverse power of *frequency*, i.e. take ramps into account. A log-log plot gives less curvature, but still ...



CALCE (Ostermann et al., ECTC 2006) fitted Engelmaier/IPC expression

$$N_F = \frac{1}{2} * K^{1/c}$$

where K is proportional to cyclic strain range, depends on alloy

$$C = c_a + c_b * \ln(1 + 360/t_d)$$

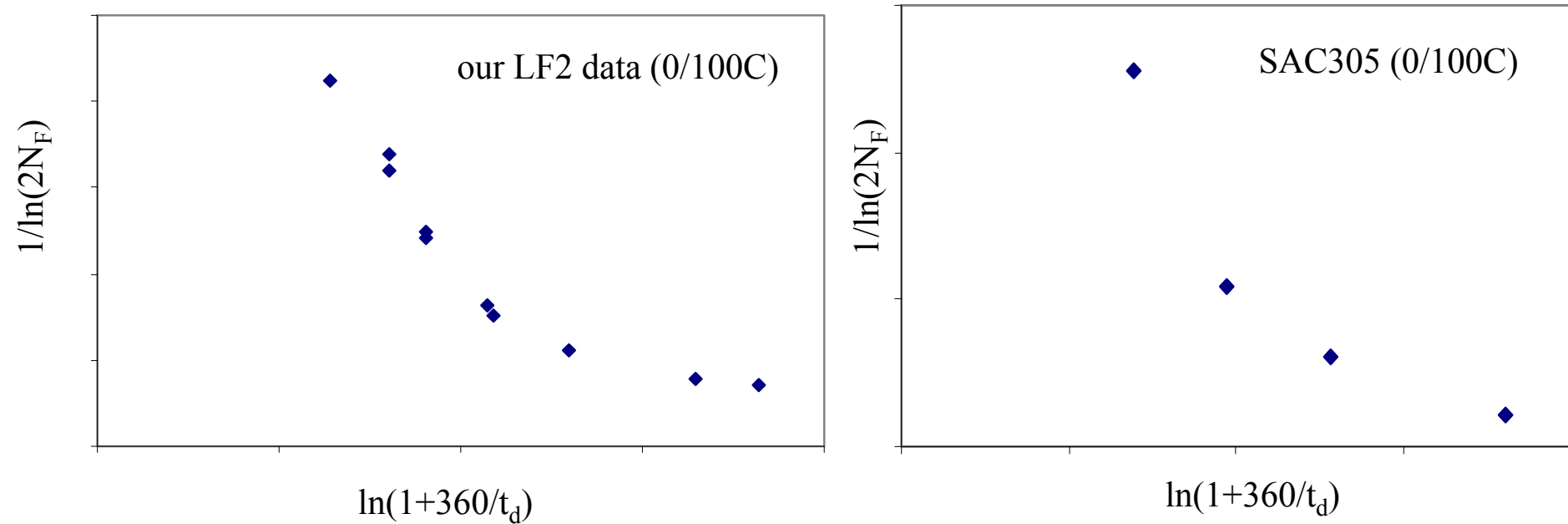
t_d is the dwell, c_b is proportional to medium temperature (!?)

However, they also only considered two t_d values for each cycling range/component. If we consider more ...

Effect of Dwell in Cycling



According to Engelmaier/IPC expression this scaling should lead to a straight line:

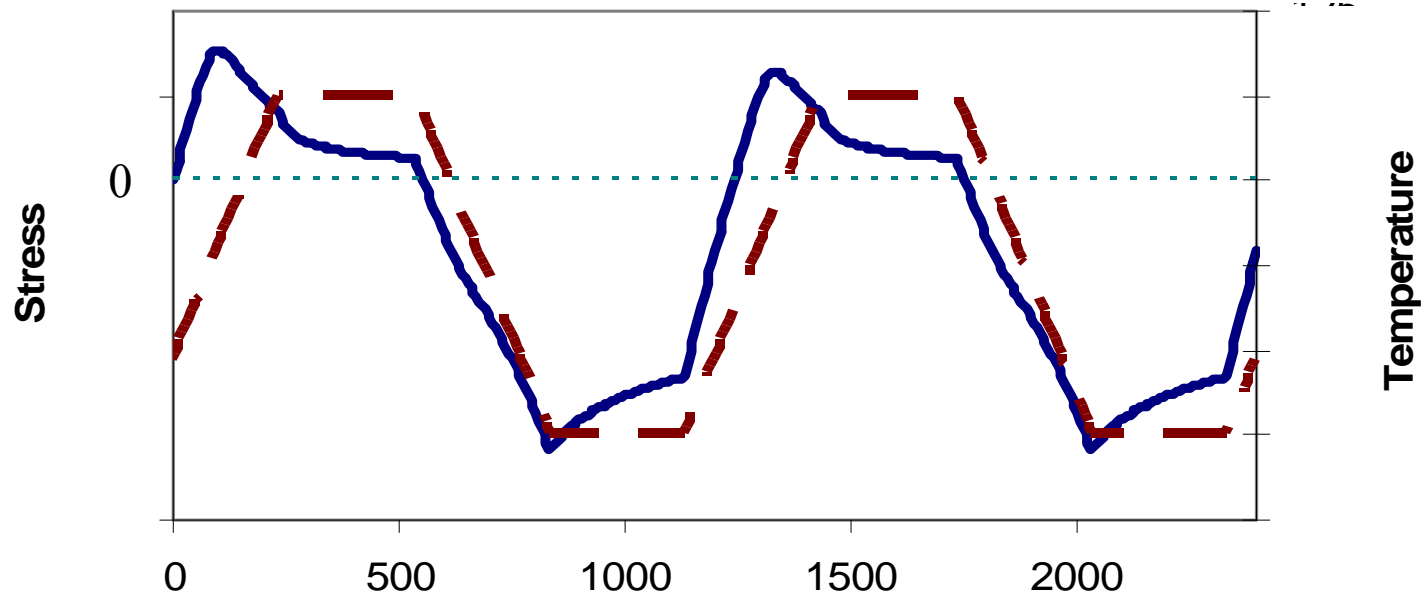


Systematic deviation from predicted dwell time dependence!

Dwell in Thermal Cycling?



Generic stress history in thermal cycling of SnPb joints (after a cycle or two they are all the same): Fatigue crack growth resistance (at same stress) drops rapidly with temperature. Dominant damage in heat up and high T dwell. Once creep has relaxed stresses sufficiently, extending dwell further doesn't matter.

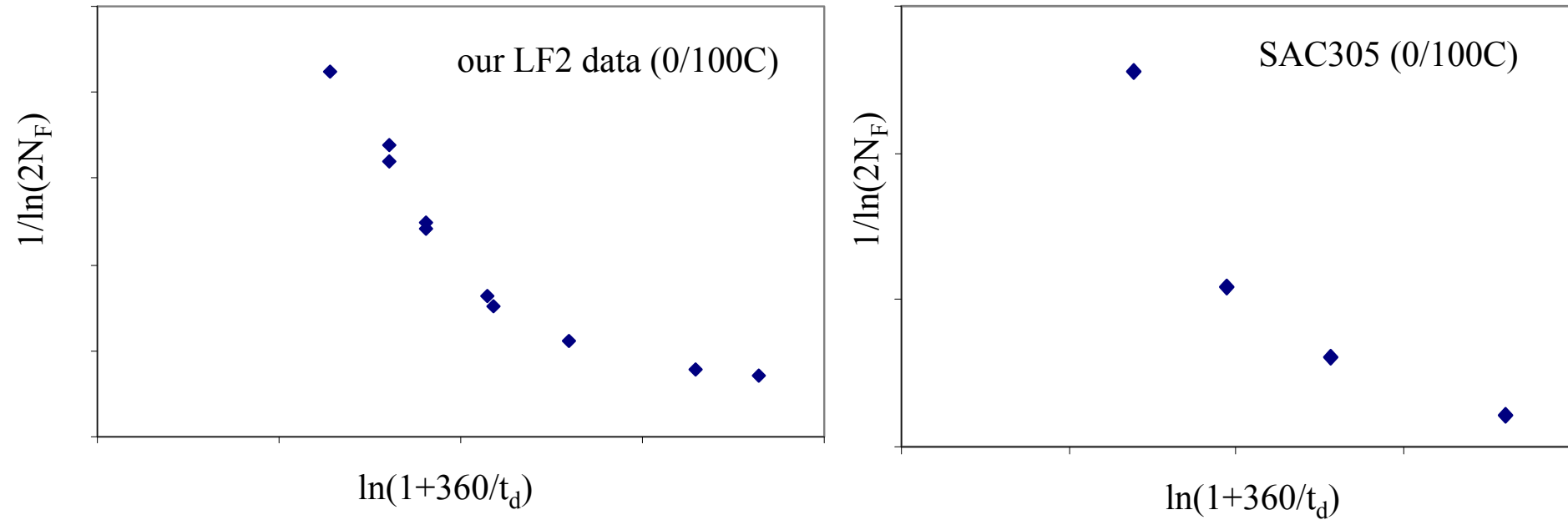


Lead free solders creep much slower, so fatigue cracks keep growing as dwell is extended (indications are also that low temperature dwell matters as well, see later).

Effect of Dwell in Cycling



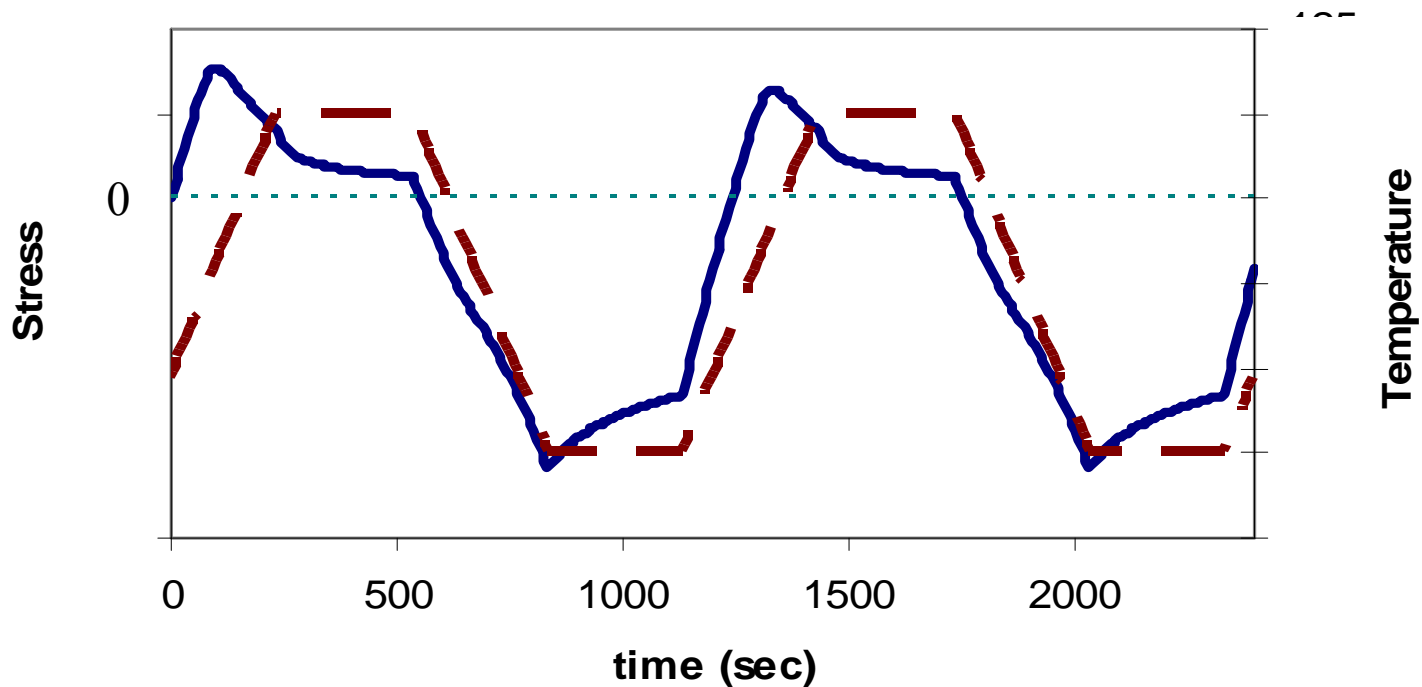
So dwell time constant (360) in Engelmaier/IPC expression *has* to be different from the one for SnPb.



Still we couldn't get a straight line with any other value for 'b' in $\ln(1+b/t_d)$ either!

Minimize (!!!) assumptions about details:

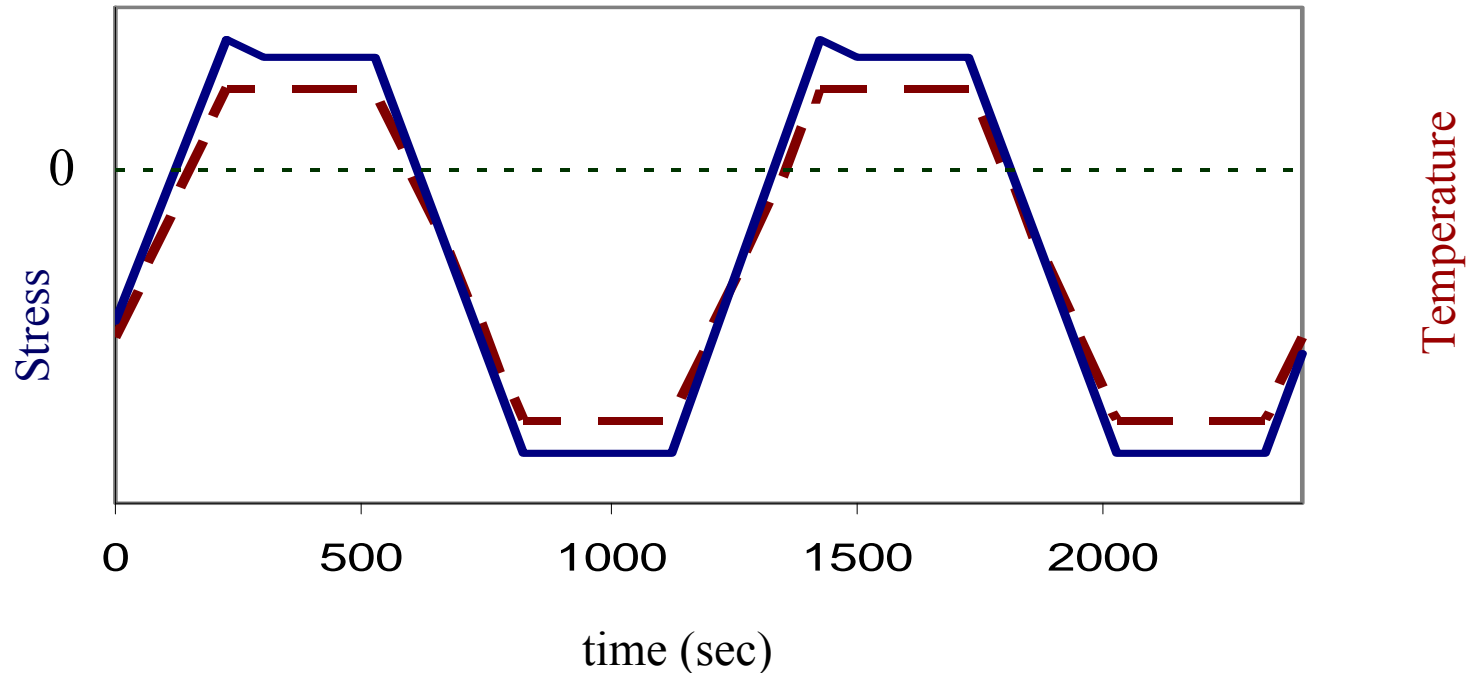
Partition damage (crack growth rate per cycle) into 4 contributions, one from each dwell and one from each ramp.



$$(dc/dN)_{\text{total}} = (dc/dN)_{\text{heat}} + (dc/dN)_{\text{hi-T}} + (dc/dN)_{\text{cool}} + (dc/dN)_{\text{lo-T}}$$

(Terms are not independent, although *ramps* may be independent of *dwells*?)

$$(dc/dN)_{\text{total}} = (dc/dN)_{\text{heat}} + (dc/dN)_{\text{hi-T}} + (dc/dN)_{\text{cool}} + (dc/dN)_{\text{lo-T}}$$

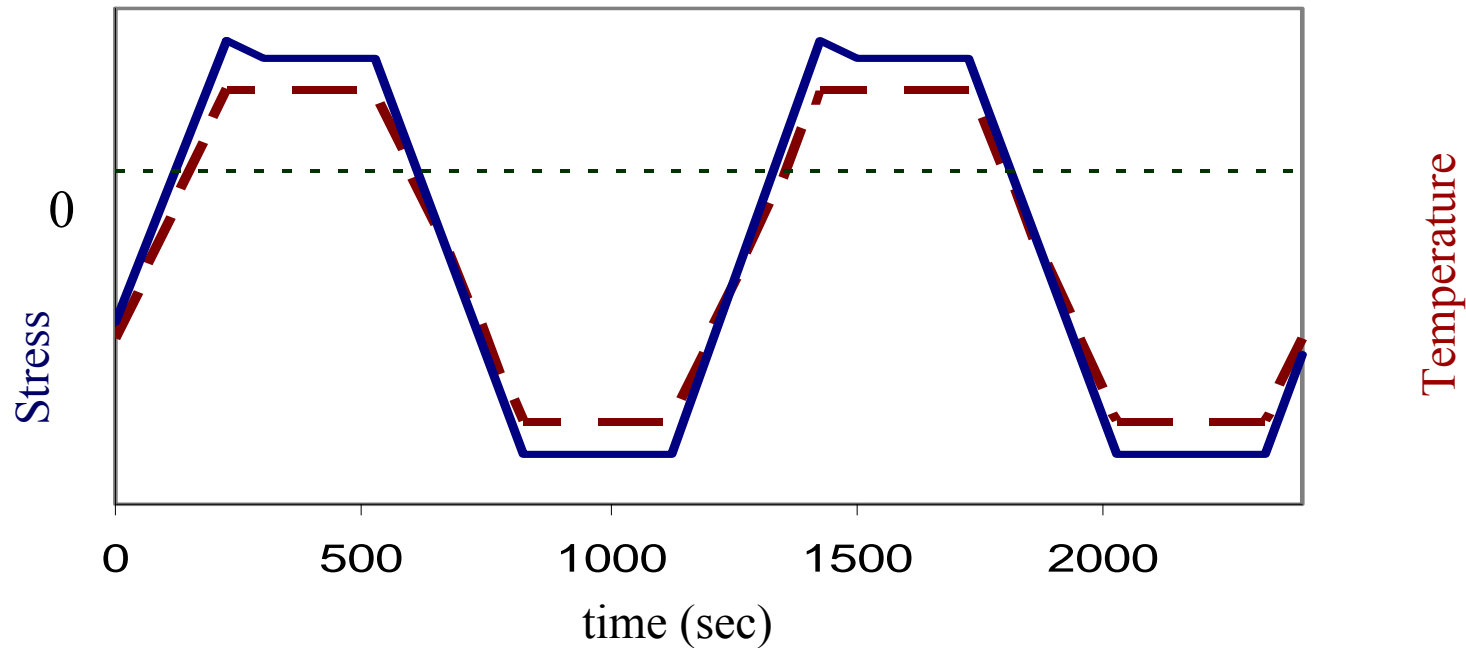


SAC stress evolution is quite different: negligible relaxation over hours at 100°C, and of course for a long time at 0°C

Our Approach (for now)



$$(dc/dN)_{\text{total}} = (dc/dN)_{\text{heat}} + (dc/dN)_{\text{hi-T}} + (dc/dN)_{\text{cool}} + (dc/dN)_{\text{lo-T}}$$

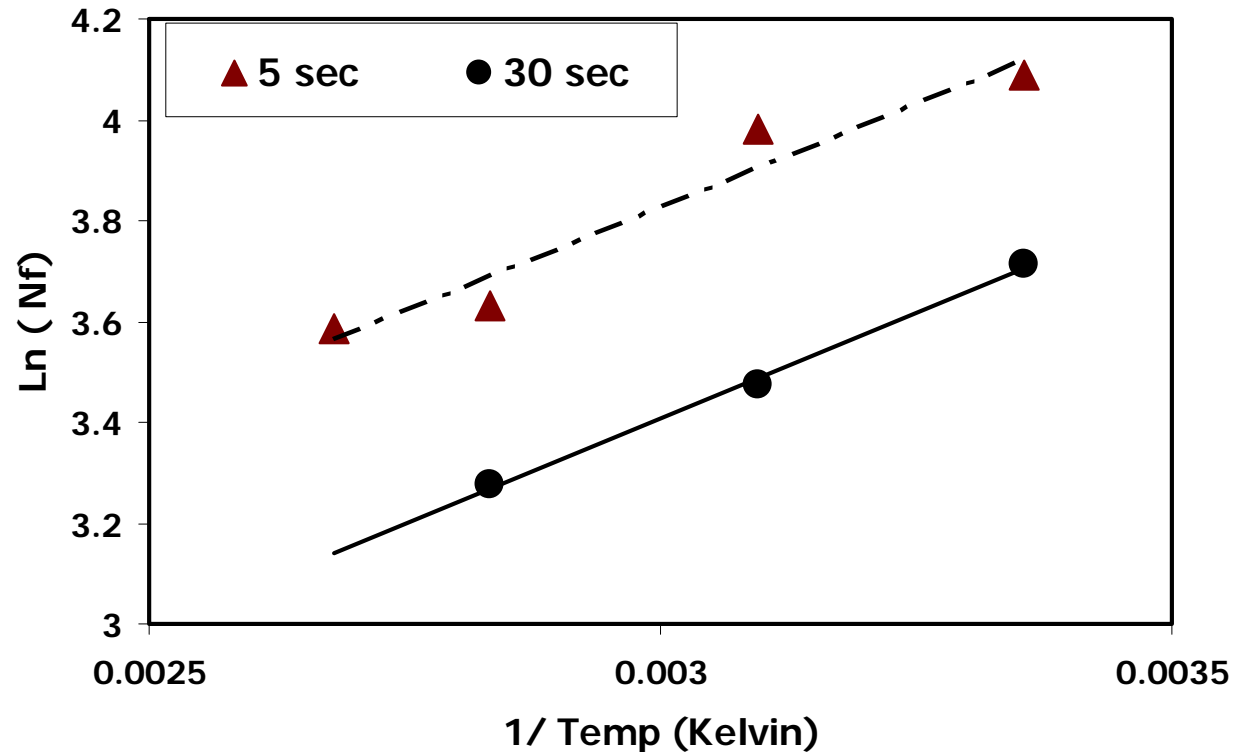


We don't have a strong argument for it, but *maybe* we can approximate crack growth *rate* in either dwell as constant (different from each other):

$$(dc/dN)_{\text{hi-T}} = a_1 * t_{d1} \text{ and } (dc/dN)_{\text{lo-T}} = a_2 * t_{d2} ?$$

We expect a_1 to depend on t_{d2} (recrystallization?) but never mind for now

Arrhenius plot for
isothermal
load controlled
fatigue life of
SAC205 with 5s
and 30s dwells



Life may not actually follow an Arrhenius dependence, but apparent 'activation energy' is essentially constant: 0.07eV (weak temperature dependence). Conversely, the initial dwell time dependence doesn't seem very temperature dependent?

So variations with dwell time at different temperatures may indeed scale?

If high and low temperature dwells are the same it follows then that
 $(dc/dN)_{\text{total}} = f(\text{ramps}) + a * t_d$.

$f(\text{ramps}) = (dc/dN)_{\text{heat}} + (dc/dN)_{\text{cool}}$ could include an initial high-stress relaxation part of dwell (but our results suggest no need to).

It does in fact appear that we can ***approximate*** by constant crack growth rate within dwell in a given cycle (not from one cycle to next):

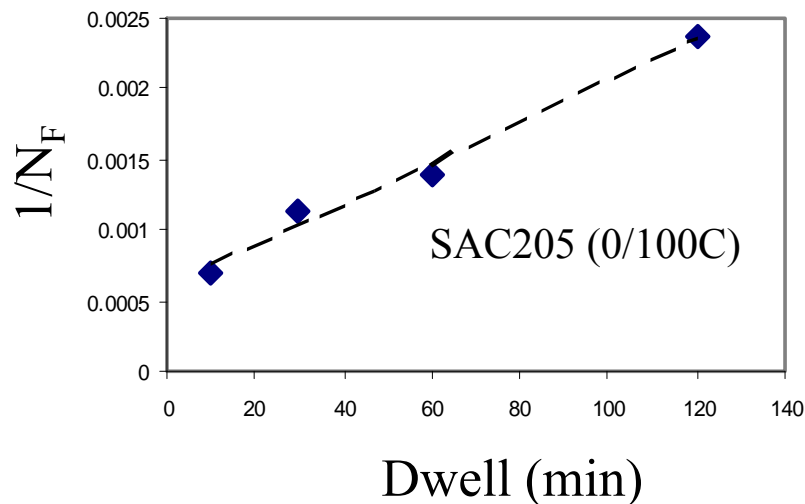
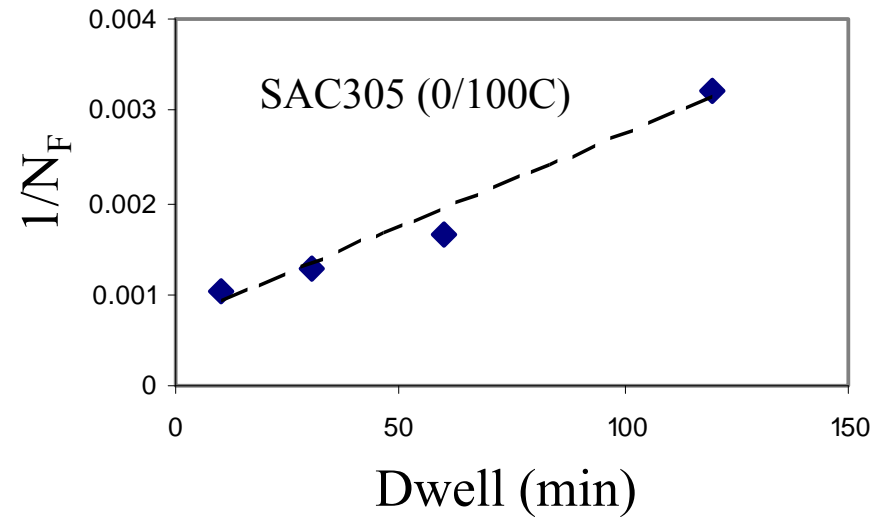
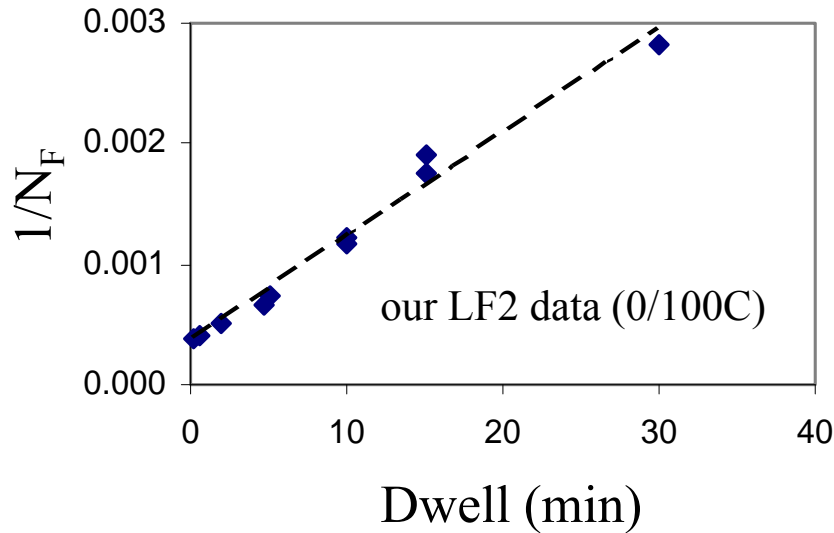
$$1/N_F = \beta + \alpha * t_{\text{dwell}}$$

where β and α are proportional to effective crack growth rates in ramps and dwell, respectively

Effect of Dwell in Cycling



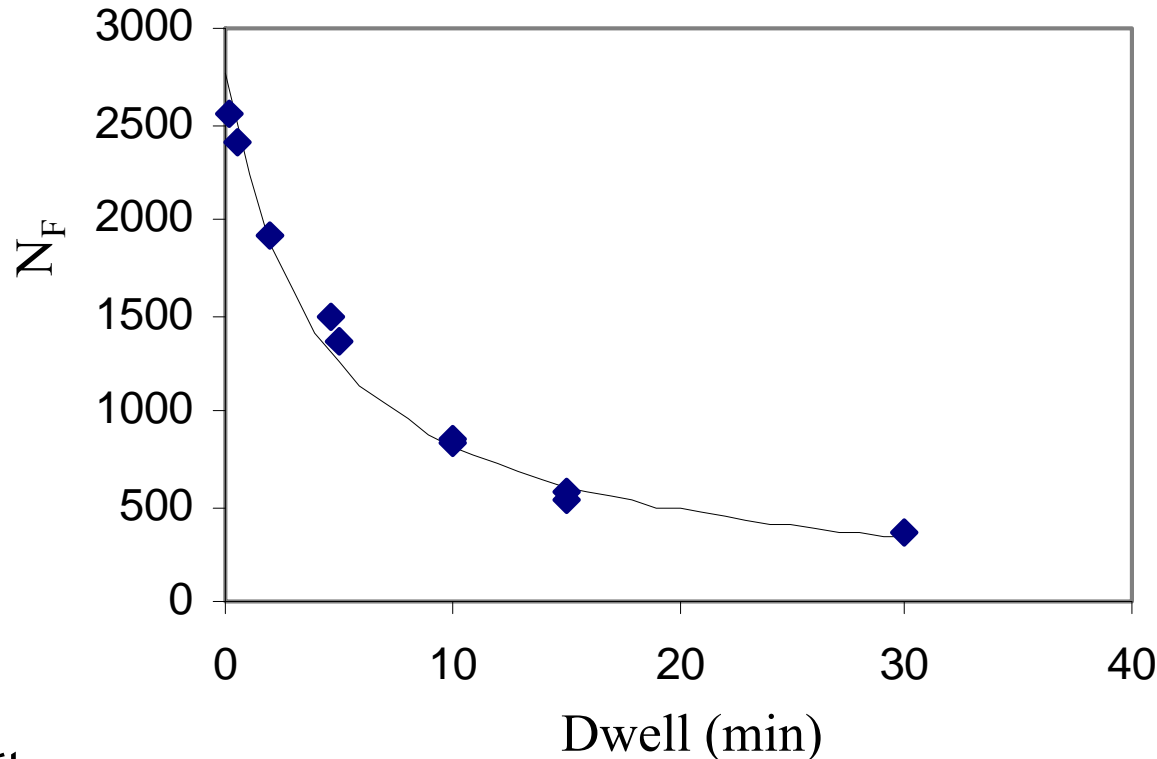
Reasonable approximation for dwell time dependence



$$1/N_F = \beta + \alpha^* t_{\text{dwell}}$$

Parameter Dependencies

So if high and low temperature dwells are the same (!!) we have a fit with only two parameters:



$$1/N_F = \beta + \alpha * t_{\text{dwell}}$$

where β and α are proportional to effective crack growth rates in ramps and dwell, respectively

Sticking with the same expression, we

$$1/N_F = \beta + \alpha^* t_{\text{dwell}}$$

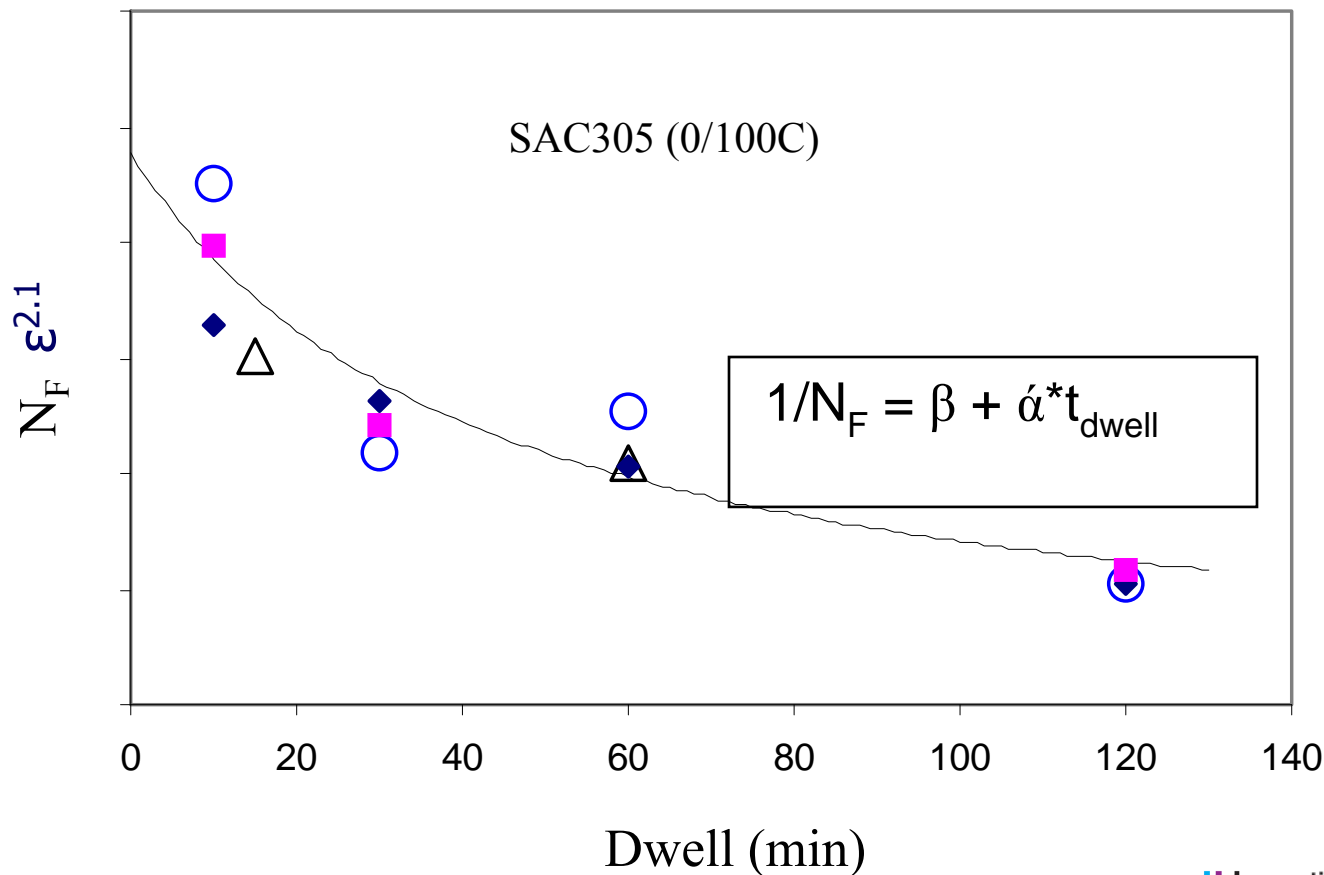
- 1) varied solder joint location (strain range, strain rate) in same cycles
- 2) varied minimum dwell temperature (ΔT , strain range) with same maximum and same ramp rates (strain rates)
- 3) varied maximum dwell temperature with same ramp rates (strain rates) and with both same and different ΔT (strain range)

for different alloys and very different solder volumes (!!).

Parameter Dependencies



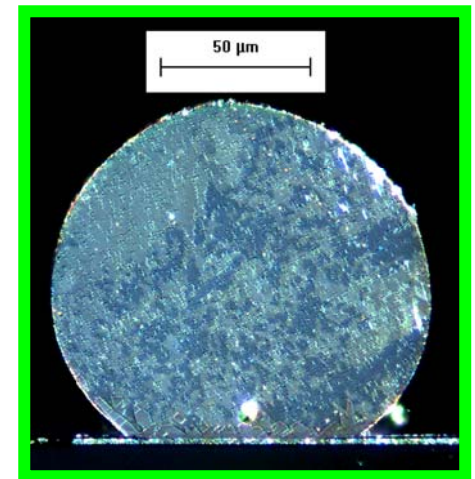
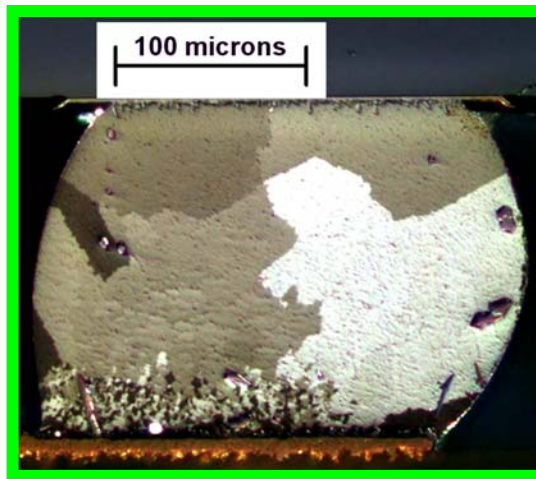
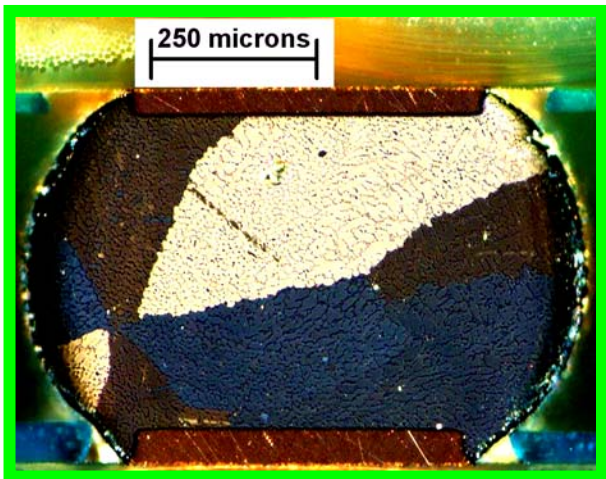
The behavior of different components with the same solder alloy and pad finishes can be scaled onto each other



However ...

Beware of joint size dependencies:

Xing, Nov. 07



Strong effect of solder volume on crack growth rate in general, but the difference between interlaced twinning and 'beach ball' structure causes further differences in behavior – so let's be careful

$1/N_F = \beta + \alpha \cdot t_{\text{dwell}}$, temperatures and ΔT constant (vary joint locations)

	1.8mm	1.4mm	1.0mm
SAC387(Zn) $\beta [10^{-4}]$	4.6	2.4	1.1
SAC105(Ni) $\beta [10^{-4}]$	5.8	3.4	1.7
SAC205 $\beta [10^{-4}]$	6.4	4.3	2.4
SAC305 $\beta [10^{-4}]$	7.7	4.2	2.1

More generally, crack growth in ramp and dwell do not show same trend with alloy: **The Ni and Zn doped alloys do best in the ramp (not because they are more ductile, SAC387(Zn) was slightly harder than SAC205), but ...**

Alloy Dependence (0/100C)



$1/N_F = \beta + \alpha \cdot t_{\text{dwell}}$, temperatures and ΔT constant (vary joint locations)

	1.8mm	1.4mm	1.0mm
SAC387(Zn) α [10^{-5} min^{-1}]	3.2	1.6	0.7
SAC105(Ni) α [10^{-5} min^{-1}]	3.6	2.1	1.1
SAC205 α [10^{-5} min^{-1}]	1.3	0.9	0.5
SAC305 α [10^{-5} min^{-1}]	1.8	1	0.5

However, the Ni and Zn doped alloys crack **faster** during the dwell(s)

Primary Parameters?



$$1/N_F = \beta + \alpha * t_{\text{dwell}}$$

(β and α proportional to effective crack growth rates in ramps and dwells)

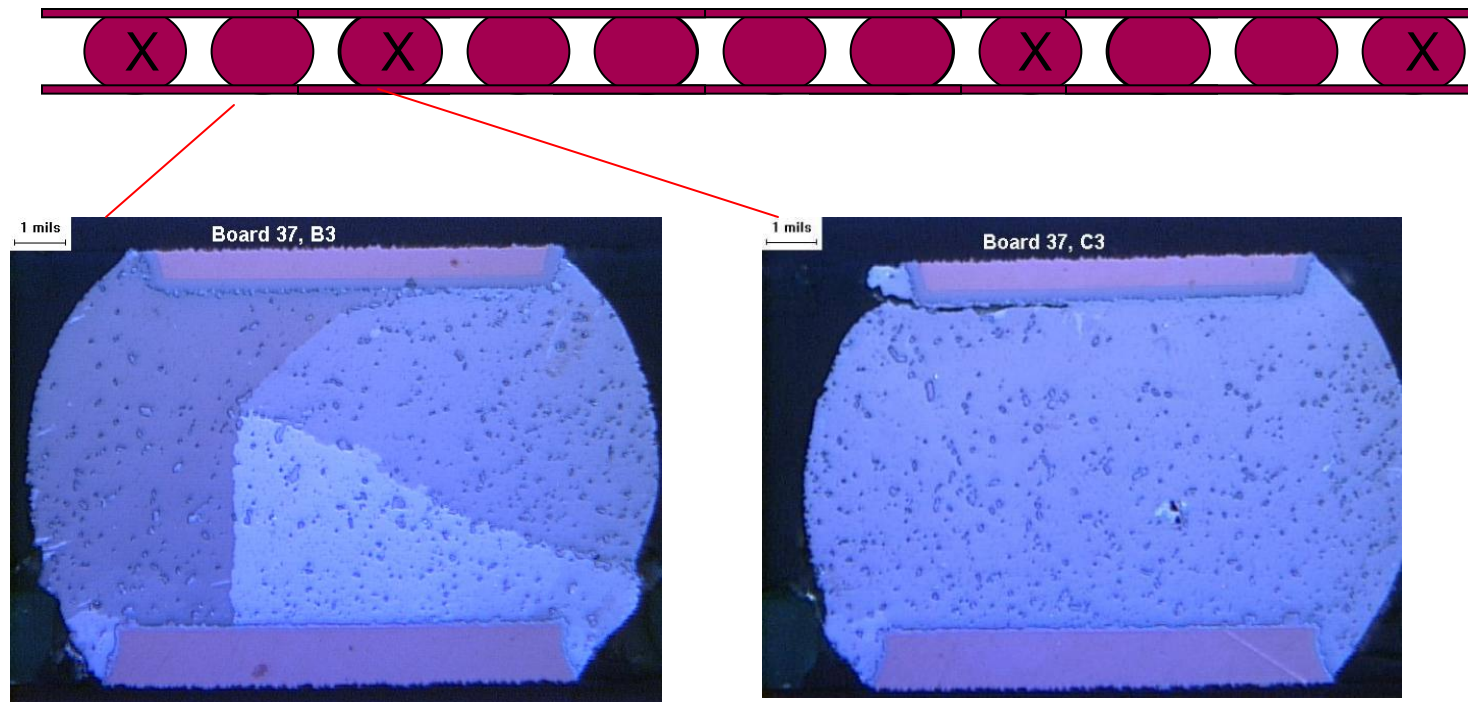
Systematics not compatible with a **direct** dependence on strain range ($\Delta\epsilon$ or ΔT). Life varies with both because

damage in the ramp is affected by
strain **rate** in ramp,
temperatures in ramp,
and time in ramp.

damage in dwell is affected by
preceding strain **rate** (but not temperatures or time in ramp),
temperature in dwell,
and time in dwell.

Alloy dependence suggests sensitivity to different materials properties in ramp and dwell

For the present components the corner joints are stressed the most and tend to fail first, but sometimes we find large cracks further in. This can be correlated with number of Sn grains and their orientations



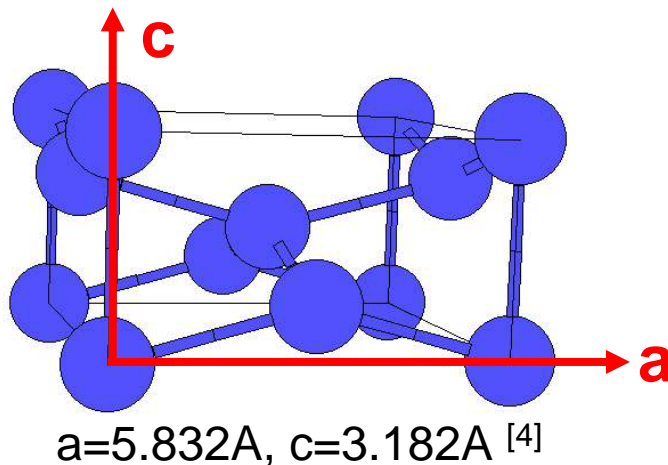
Fastest crack growth in single grained joints with particular orientation (the strongest)

Sn is strongly anisotropic



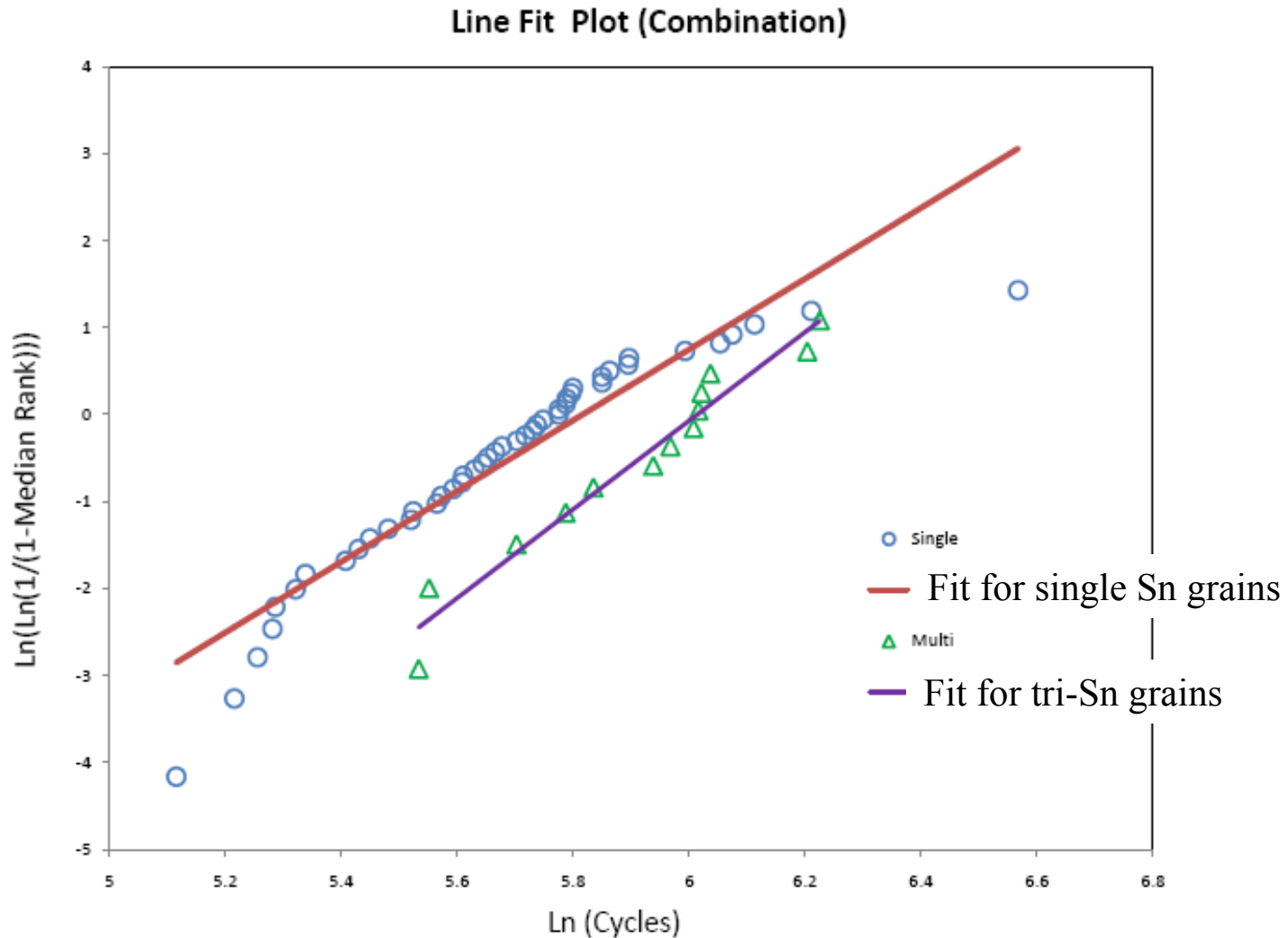
Particularly rapid cracking if c-axis is parallel to pad surface & loading

Direction < >	Thermal Expansion Coefficient ($10^{-6}/^{\circ}\text{C}$)	Young's Modulus (Gpa)
	23 ^[1]	41 ^[1]
c <001>	30.5 ^[2]	67.6 ^[3]
a <100>	15.4 ^[2]	23.6 ^[3]



- Sn has a body centered tetragonal unit cell: $a = b = 5.83 \text{ \AA}$, $c = 3.18 \text{ \AA}$
- **c** direction expands and shrinks more with a temperature change
- **c** direction is more stiff while **a** direction is more compliant

Single and multi grain joints have different failure distributions



Accelerated testing does, by nature, run the risk of **missing effects of long term aging.**

The ***most likely*** way of missing something may be RT preconditioning followed by thermal cycling with maximum temperature above operating temperature of concern:

- Dissolution at high temperature may 'wash out' effects of aging on precipitate distributions

- Temperature and strain induced ripening of precipitates and IMC layers may 'drown' effects.

Effects on vibration, drop, bending, ... generally larger

We don't know acceleration factors for aging.

How do we assess reliability of 20 year old joints without waiting 20 years?

Thermal cycling, vibration, shock, bending?

Joints may fail in solder, IMC bond, or by pad cratering.

SAC joints on Cu or Ni/Au pads: $(\text{Cu}, \text{Ni}, \text{Au})_6\text{Sn}_5/(\text{Ni}, \text{Cu})_3\text{Sn}_4$, $(\text{Cu}, \text{Ni})_6\text{Sn}_5/\text{Cu}_3\text{Sn}$, ..., IMC structures on solder pads grow thicker, weakening (perhaps degrade). Quantitative acceleration is non-trivial, and solder properties is a factor too.

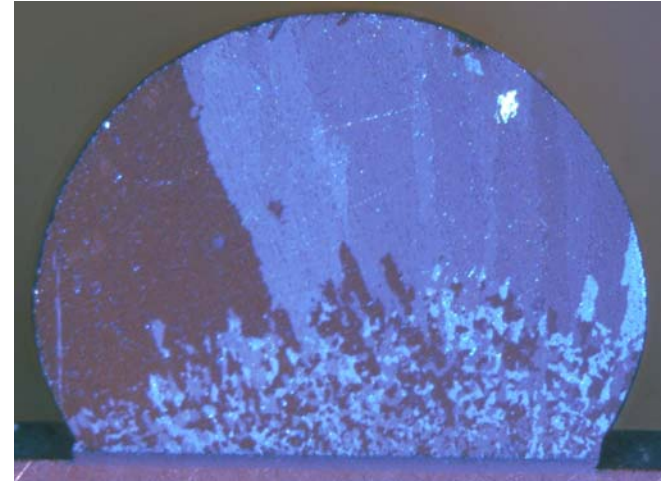
A little aging may actually improve cratering resistance (further curing laminate), but primary concern is degradation in humidity. Quantitative acceleration is non-trivial, and solder properties is a factor too.

Solder properties affect failure by cratering, IMC cracking or solder cracks.

Properties change over time – faster in cyclic loading.

The state of aging is characterized by the solder joint microstructure, which determines its mechanical properties and reliability.

1. Solder deformation and failure depend on solder microstructure, which depends, in turn, on how the solder is processed.
2. Common solders are microstructurally unstable. Due to its high homologous temperature, as the solder joint is aged, thermally cycled or deformed, its microstructure evolves so that its mechanical behavior changes with time.

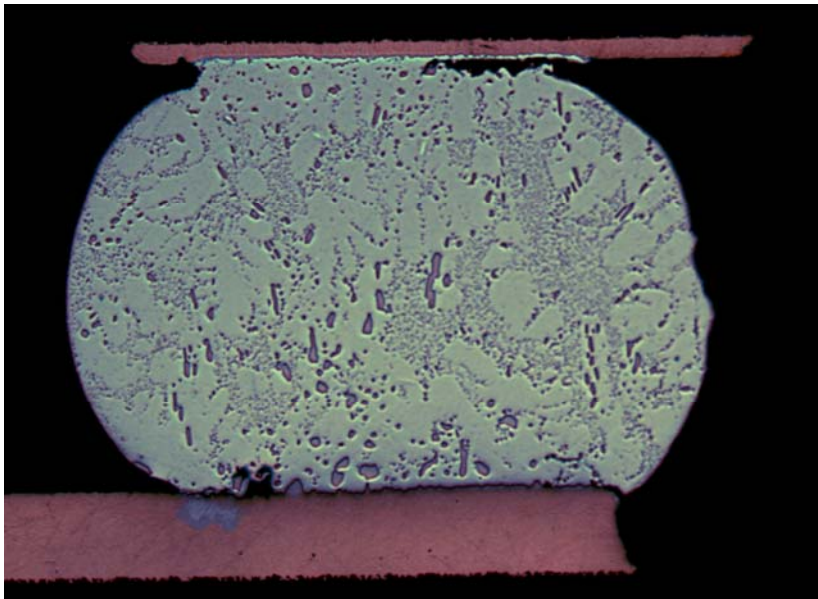
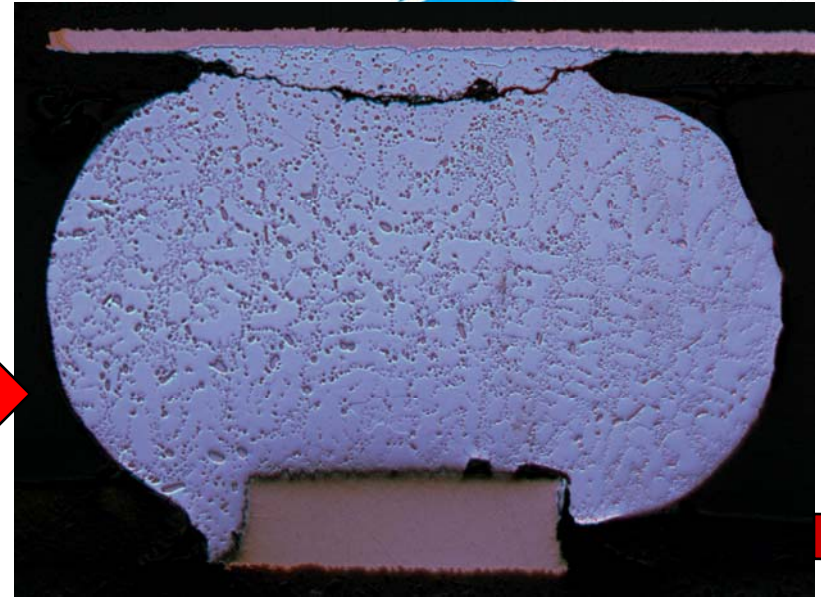
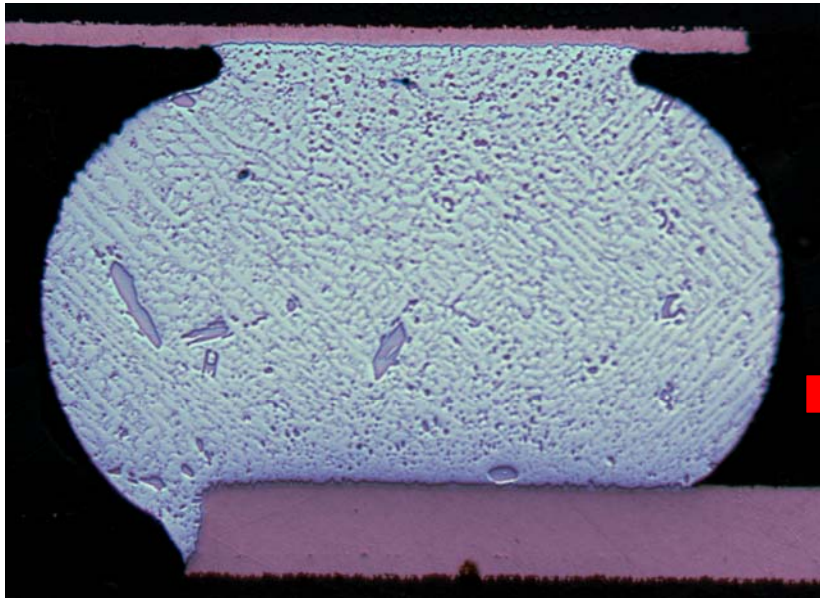


SnAgCu solder joints on Cu and/or Ni/Au pads:

- Sn grains may grow

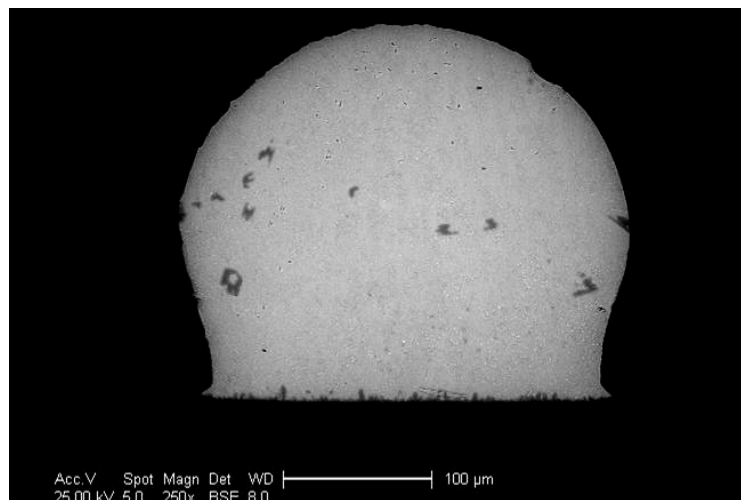
- Dendrites may coarsen

- $(\text{Cu}, \text{Ni}, \text{Au})_6\text{Sn}_5$, Ag_3Sn , $(\text{Au}, \text{Ni})\text{Sn}_4$ precipitates ripen, softening solder



Dendritic coarsening in thermal cycling (less in corresponding aging)

Cu_6Sn_5 Ripening at 125°C



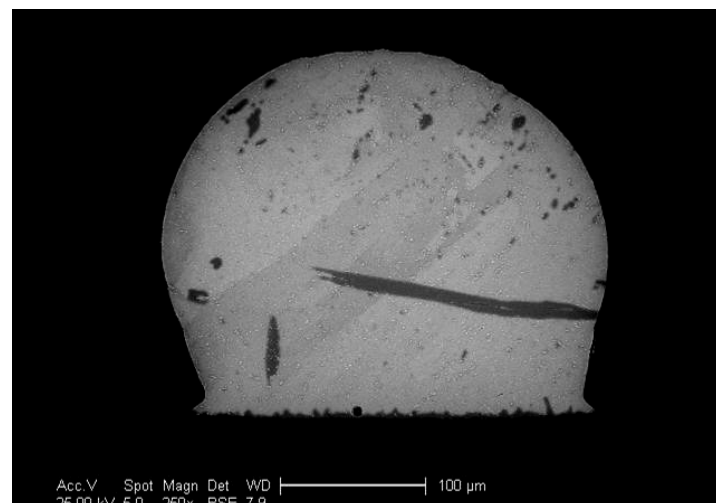
After reflow



3 days

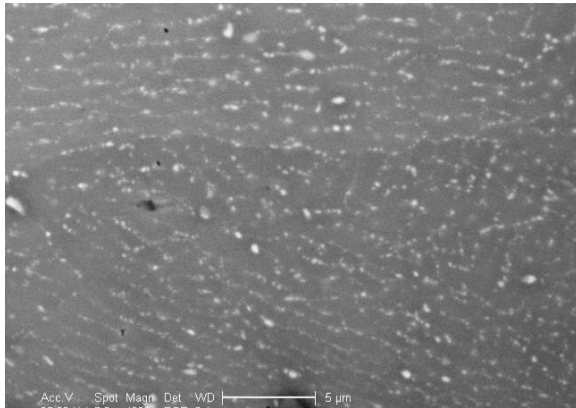


4 weeks

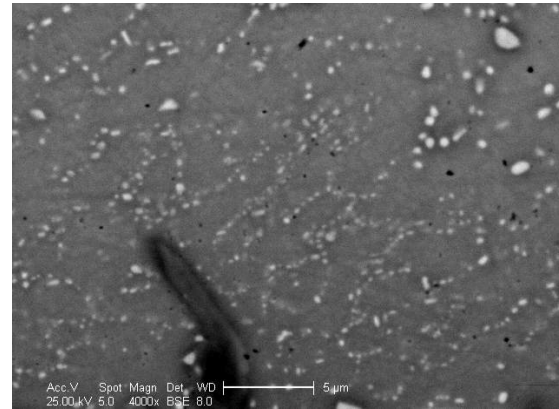


14 weeks

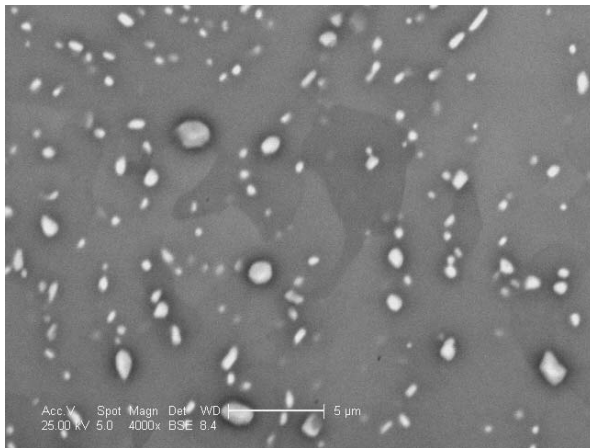
Ag₃Sn Ripening at 125°C



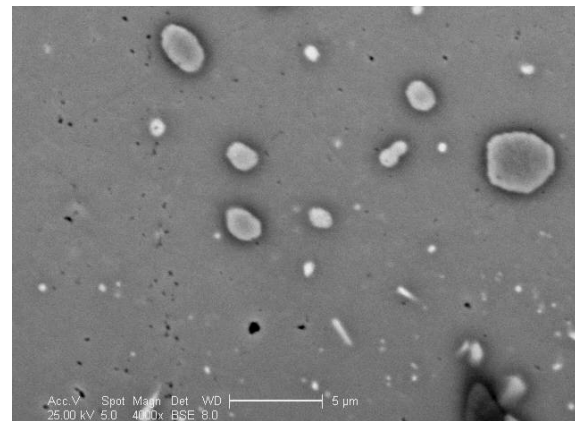
After reflow



3 days



4 weeks

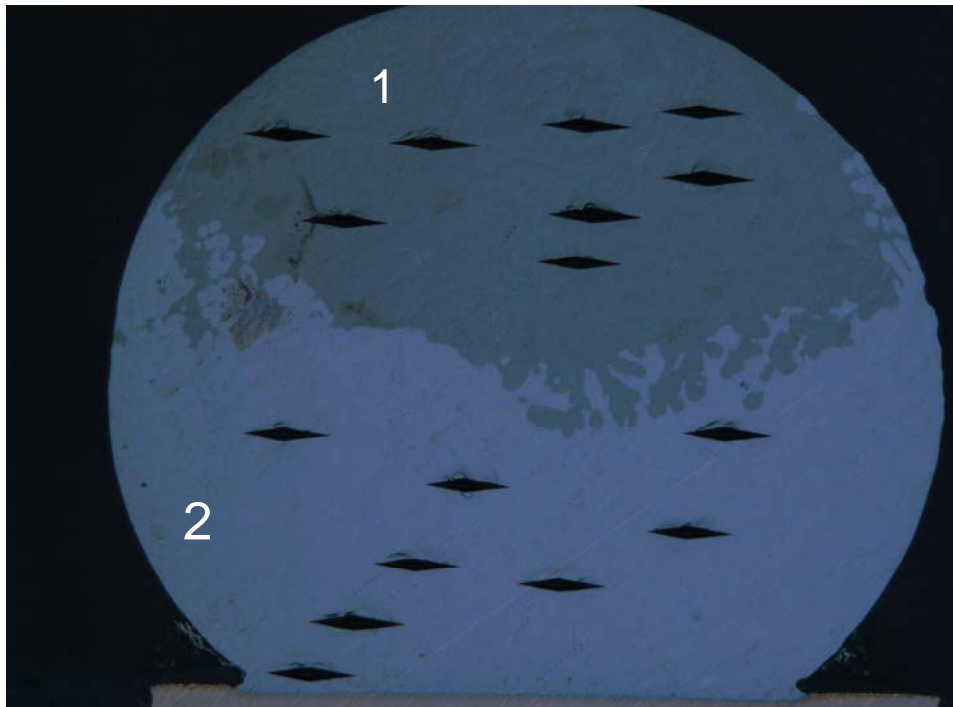


14 weeks

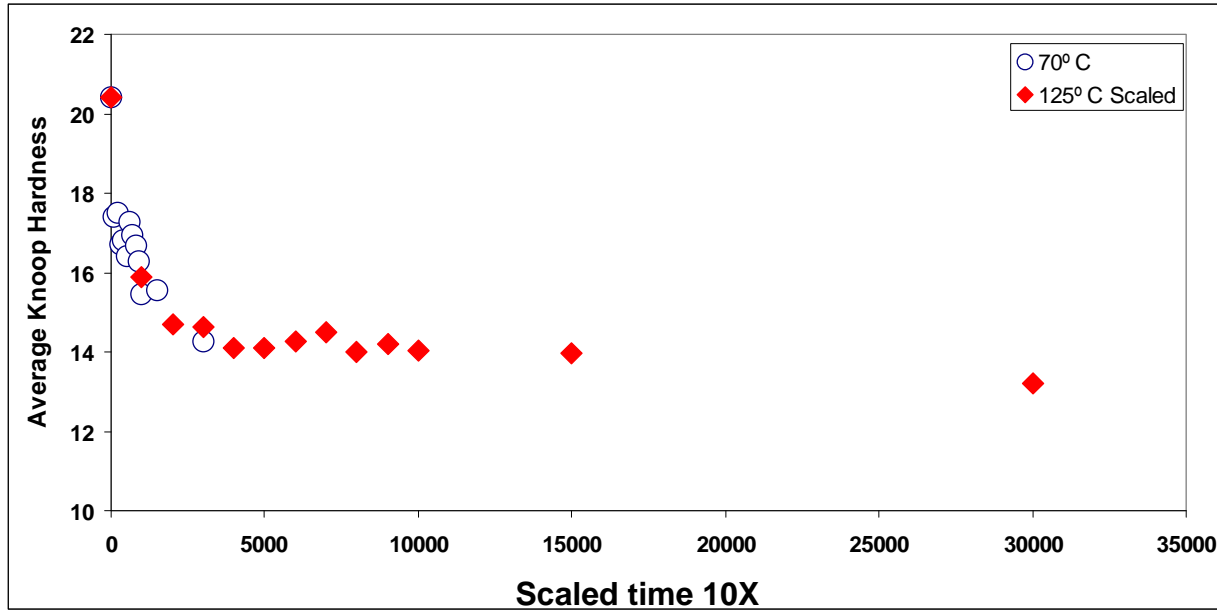
Acceleration factors: which of these looks like the image of a joint after, say, 1 year at 70C? (the 4 week one does)

As expected, not all microstructure changes vary at same rate with temperature.

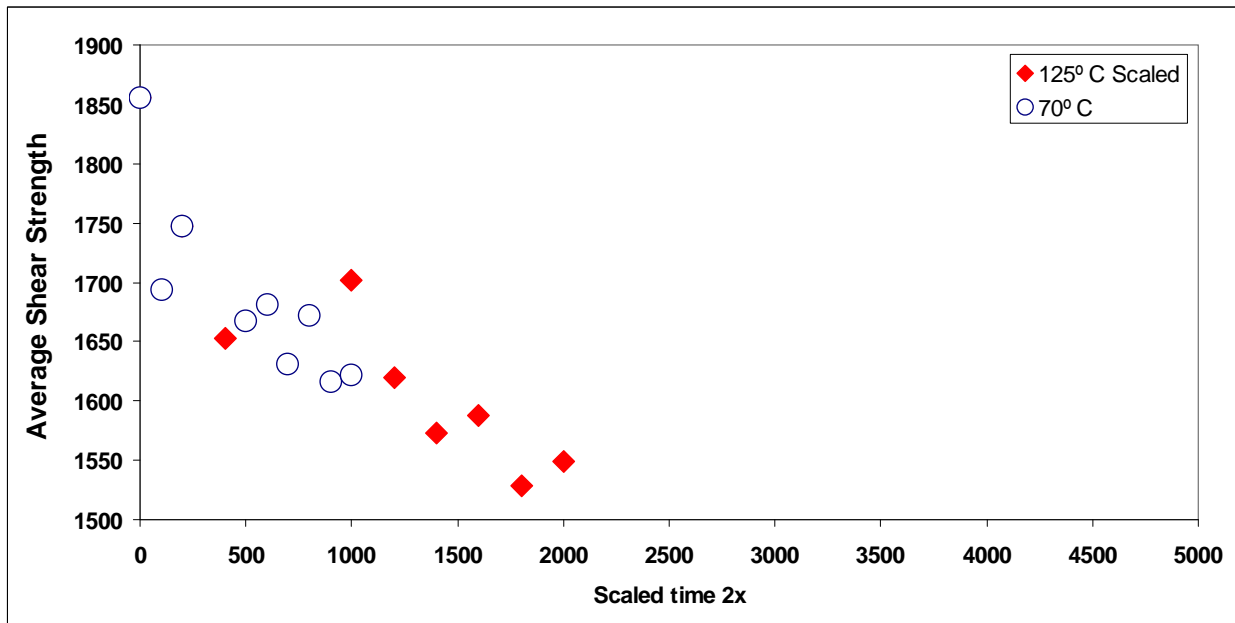
The same is true for mechanical properties (microhardness, creep, strength, fatigue resistance, ...)



Aging of SAC305



Subsequent RT hardness drops 10x faster, strength 2x faster, in aging at 125°C than at 70°C



SAC305:

Microhardness reduction, creep, and Ag_3Sn ripening have similar activation energies

Strength reduction activation energy much lower

Activation energy for load controlled fatigue life reduction higher than for strength

SAC-alloys:

Microhardness activation energy decreases with increasing Ag content

.....

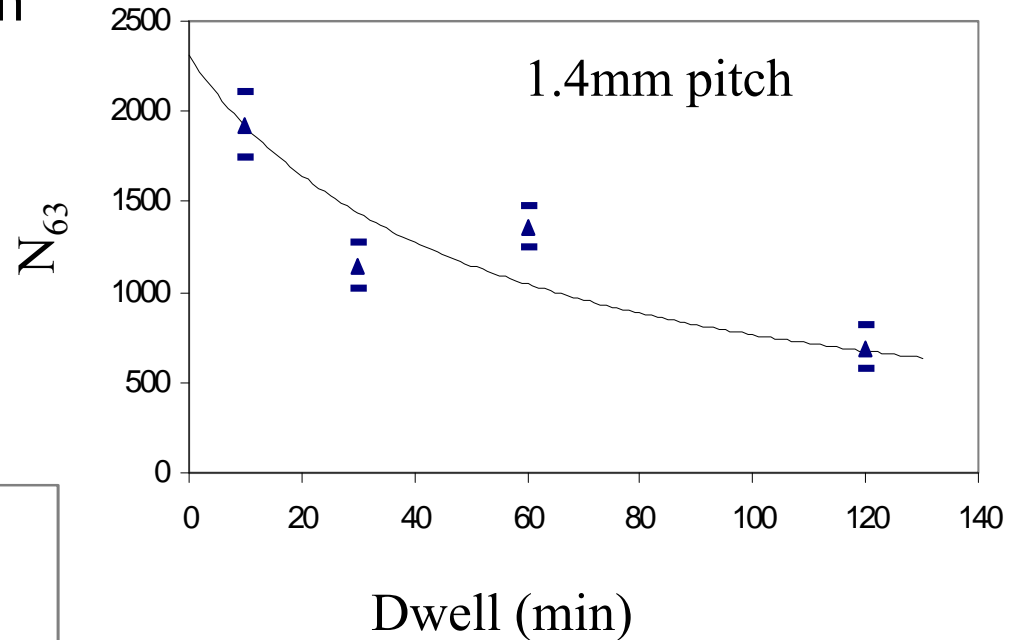
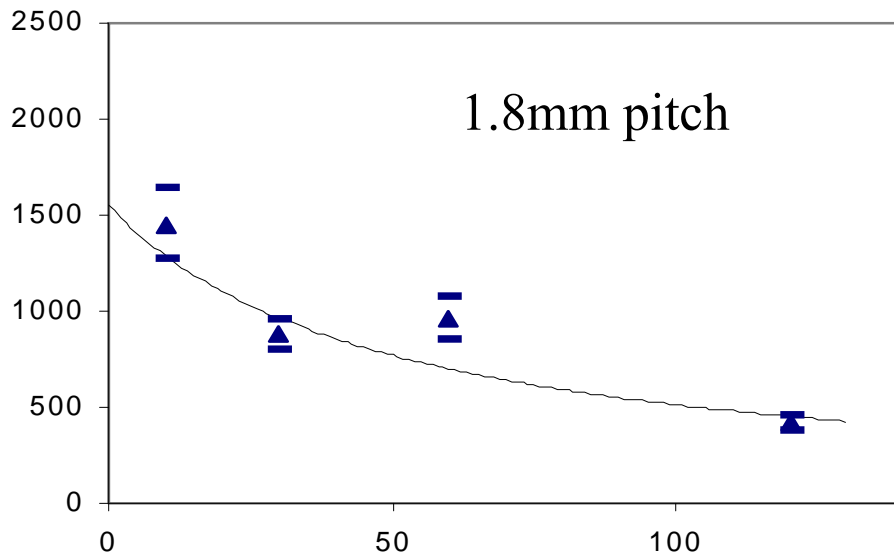
You cannot establish the *same* microstructure faster by heating – we need to be much smarter about acceleration

'Imperfection'



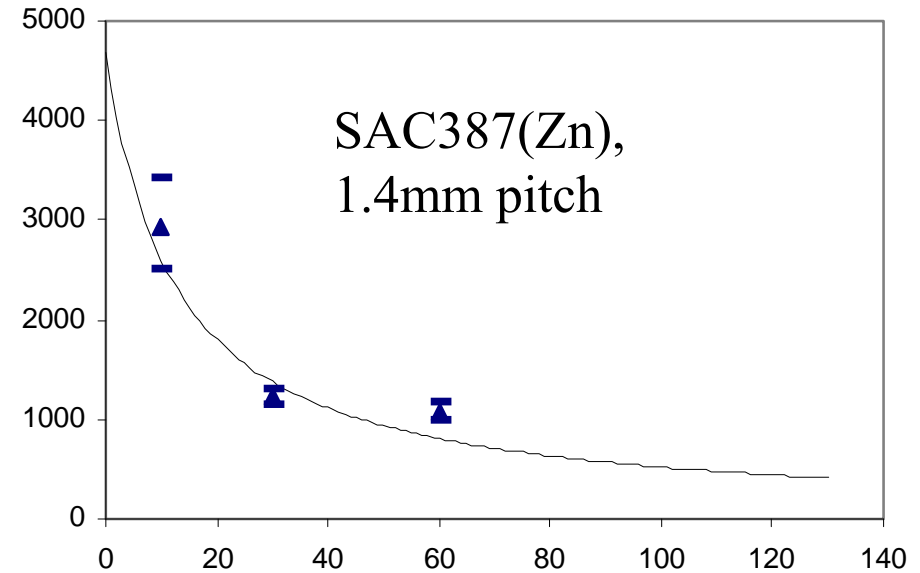
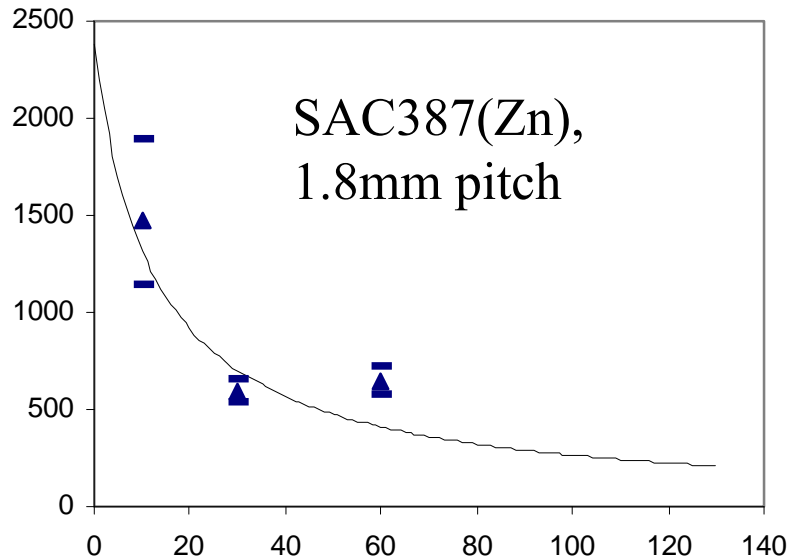
Dwell time dependencies for different alloys and strains (in same cycling) show reproducible 'bump' in same dwell time range

SAC205



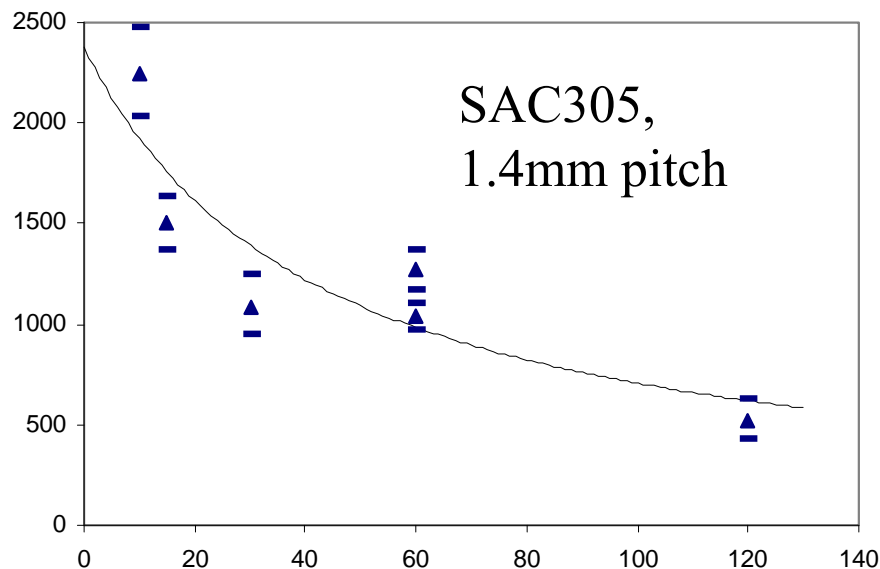
Slightly *more* cycles to failure with 60 minute dwell than with 30.

'Imperfection'



Dwell (min)

The effect is small but real: The location of 'bump' in life vs. dwell (~60min) seems independent of strain, alloy, ... in 0/100C



Just an imperfection?



Aside from a limited 'bump' in the curve we seem to be able to describe dwell time dependence, so should we worry?

Yes: Preliminary indications are that short dwells (and high strains) may lead to different microstructures and materials properties *during the test* than longer dwells (and lower strains) !!!



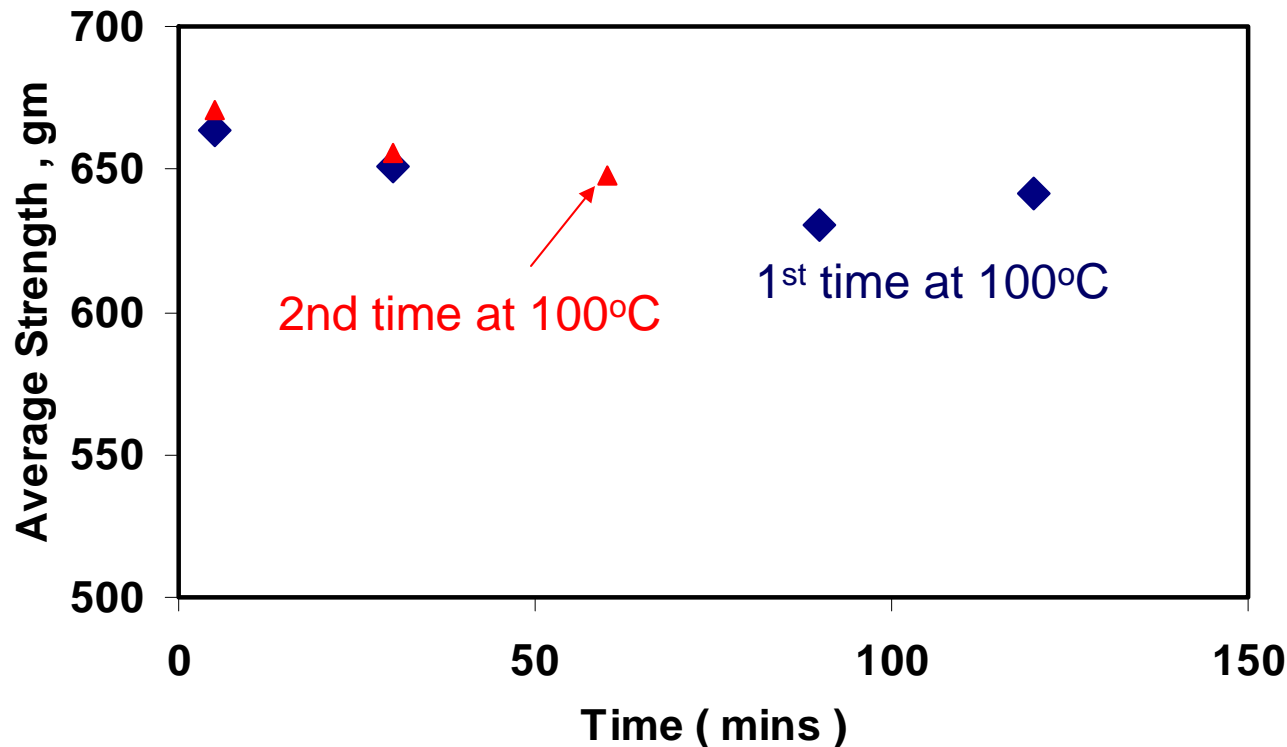
In general:

Properties vary with time at temperature,
and

crack growth mechanism depends on dwell (cracking with and without recrystallization is not the same)

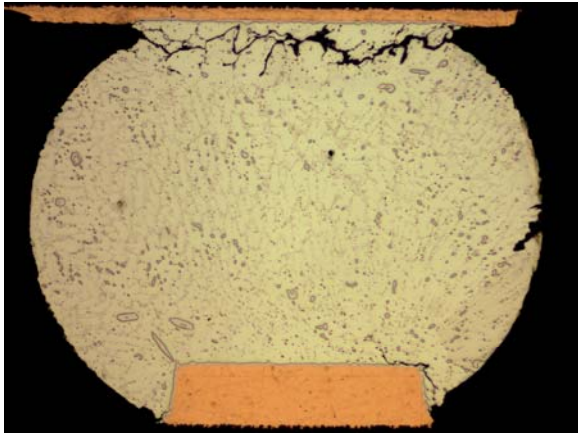
- Solder properties (reflected here in shear strength) vary during single high temperature dwell
- This is **not** cumulative effect of high temperature exposure

Strength Test @ 100 C

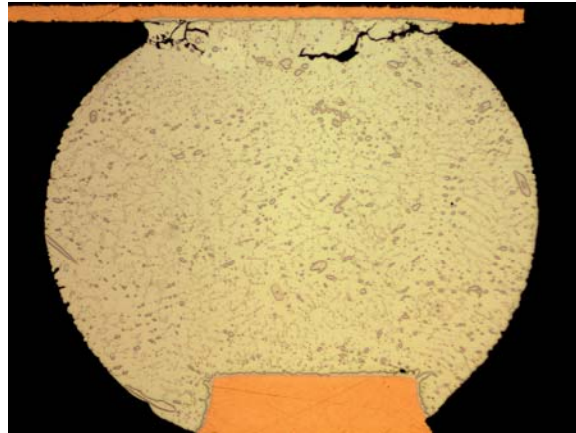


Isothermal shear strength vs. time at 100°C. Second time after 120 minutes at 100°C and 1 hour at RT.

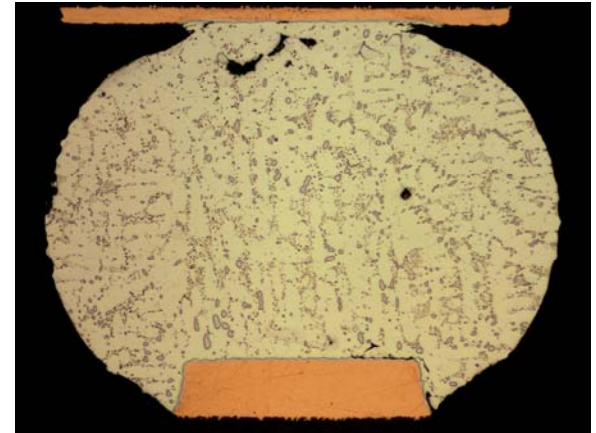
Cracking/recrystallization for different dwells



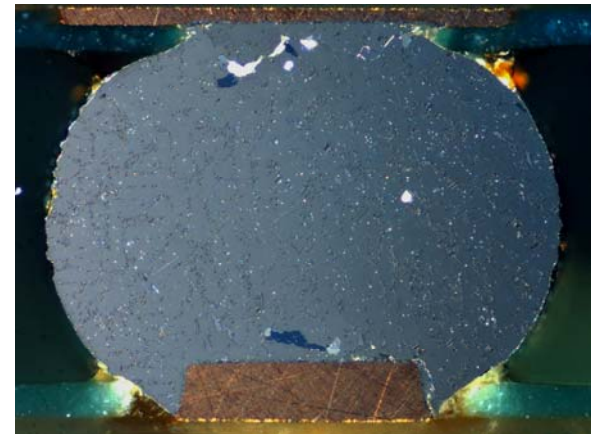
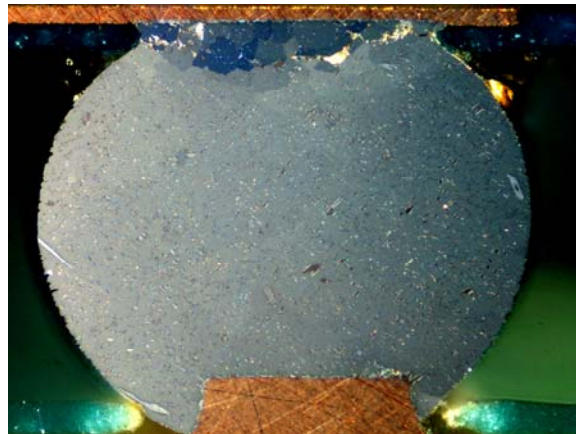
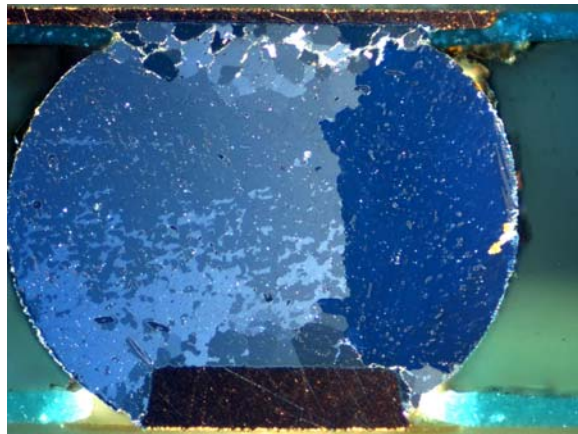
DT: 10min. 833hr.@100C



DT: 30min. 300hr.@100C

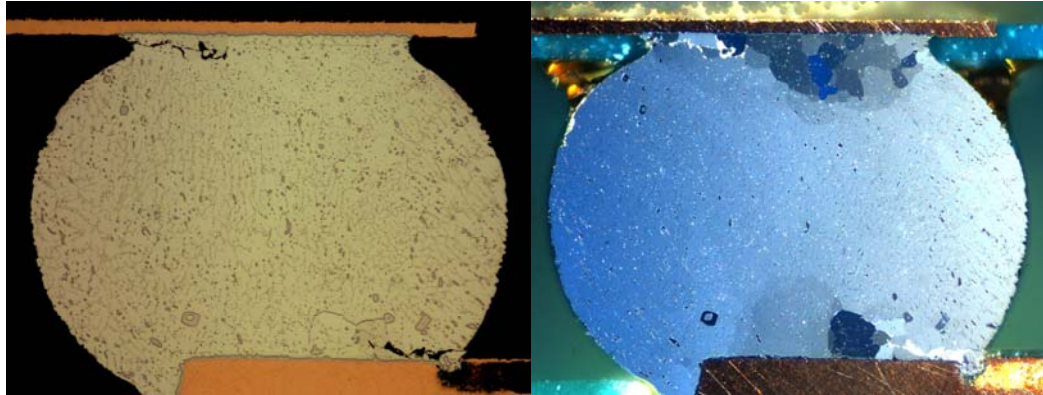


DT: 60min. 2200hr.@100C



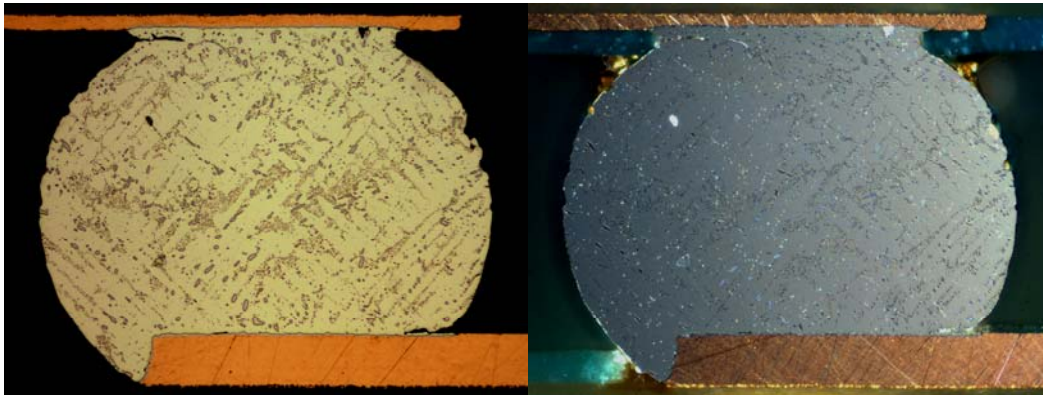
- It seems that a longer dwell time may allow more recovery in each cycle – dislocation annihilation before recrystallization

Cracking w/o recrystallization for 1+hr dwells



DT: 30 min.

~1300hr.@100C



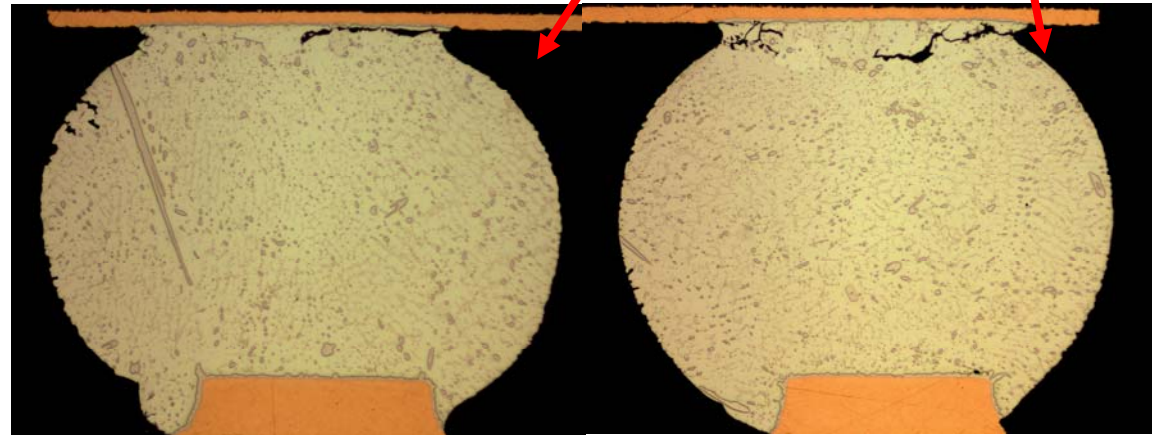
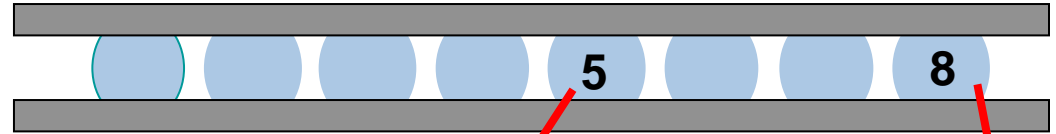
DT: 60 min.

~2200hr.@100C

For the present relatively high strains!

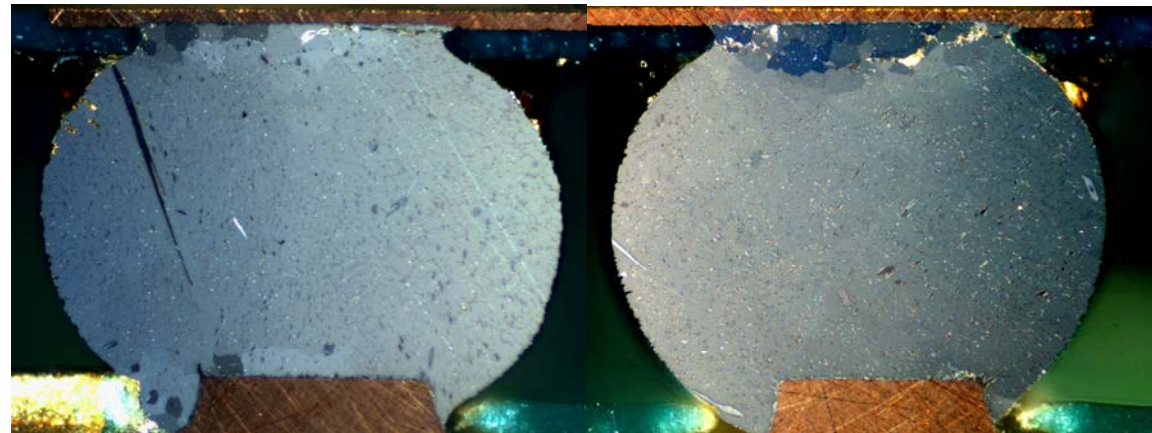
It is not as simple as that:

in lower stress locations
even a 30 minute dwell
seemed to be enough for
recovery

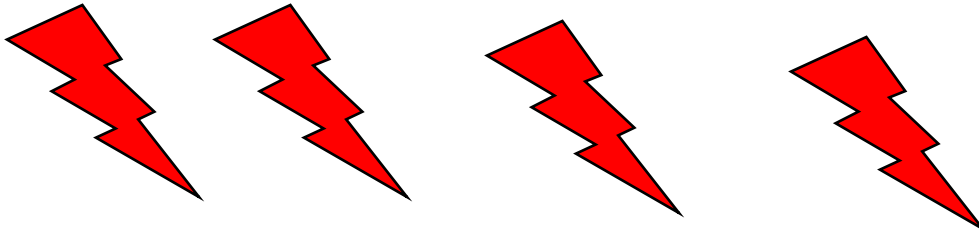


DT: 30 min.

~1300hr. @100C



So unless properly accounted for **accelerated test may involve cracking of a different material (state) than life in service!!**



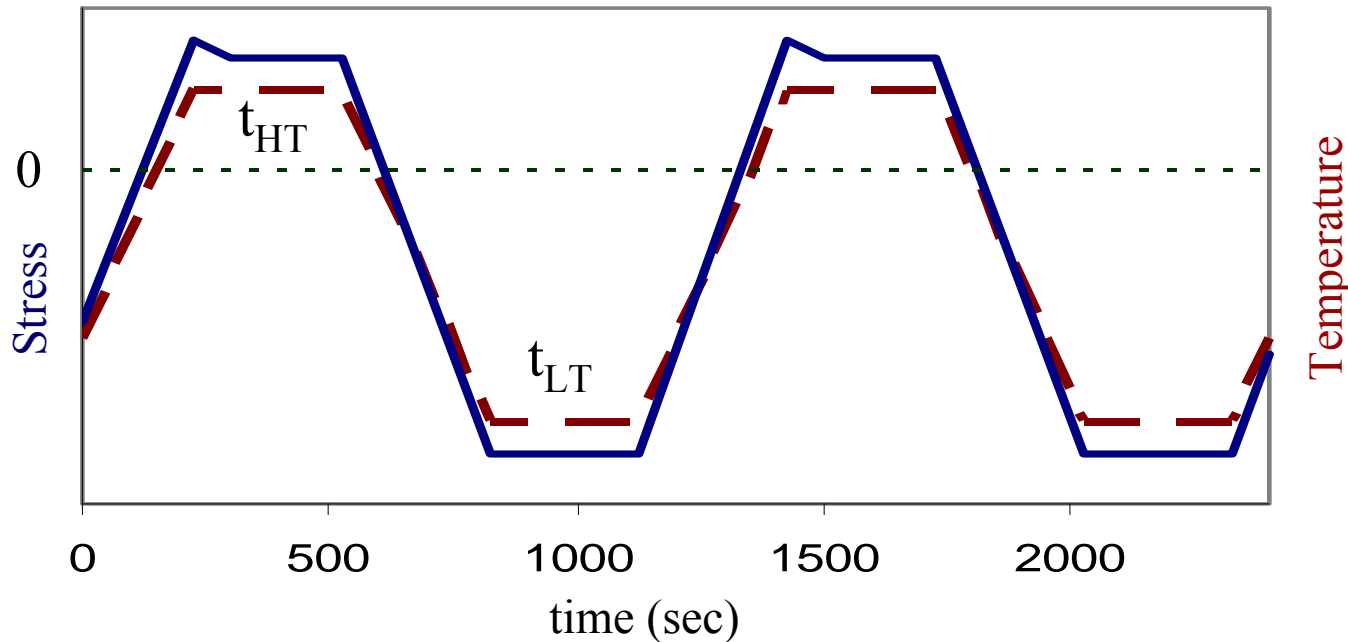
Finally, remember that previous reliability assessment models and expressions were developed for SnPb where the dominant damage occurs near the maximum temperature.

This is not the case for Pb free:

Low Temperature Dwell?



Meilunas, 2004: Vary high and low temperature dwells, t_{HT} and t_{LT}



0/100C shock
very high ramp rates:

Damage depends on
both dwells!

t_{HT} (sec)	t_{LT} (sec)	N_F (cycles)
30	30	2414
900	30	1320
30	900	1183
900	900	524

Does damage evolution continue while the computer is off?

Clearly not over long times, but consequences yet to be assessed.

Is it more complex:

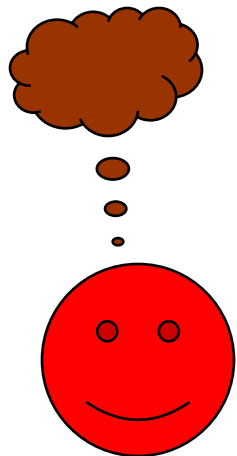
dislocations pile up under constant load at low temperature, recrystallization may occur at high temperature?

If so, *combination* of dwells may matter (interactions)?

There are many things we don't know yet, in terms of both limitations (notably extrapolations towards lower strains/longer life) and parameter dependencies.

In fact, there are so many dependencies and potential interactions to address that we haven't had time to get much statistics on the trends we do see.

(Nevertheless, our preliminary approach beats current alternatives)



borgesen@uic.com
607-768-2132
pborgese@binghamton.edu