National Aeronautics and Space Administration

The effect of prior exposures on the notched fatigue behavior of disk superalloy ME3

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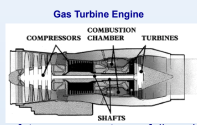
Motivation: Environmental attack has the potential to limit turbine disk durability [1,2], particularly in next generation engines which will run hotter; there is a need to understand better oxidation at potential service conditions and develop models that link microstructure to fatigue response.

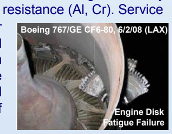
Introduction

Gas Turbine Engine



More efficient gas turbine engine designs will require higher operating temperatures. Turbine disks are regarded as critical flight safety components; a failure is a serious hazard. Low cycle fatigue is an important design criteria for turbine disks. Powder metallurgy alloys, like ME3, have led to major improvements in temperature performance through refractory additions (e.g. Mo,W) at the expense of environmental resistance (Al, Cr). Service conditions for aerospace disks can produce major cycle periods extending from minutes to hours and days with total service times exceeding 1,000 hours in aerospace applications. Some of the effects of service can be captured by extended exposures at elevated temperature prior to LCF testing [3,4]. Some details of the work presented here have been published [5].

Disk Rim	Current → 650 °C Long-range goal → 800 °C



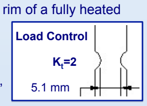
Experimental approach

Oxidation can reduce fatigue life in disk alloys above 650°C by accelerated crack initiation and growth at defects; however, it is not well-studied at 650 °C - 800°C:

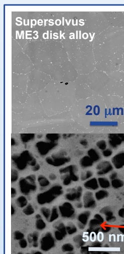
Part I
- Identify microstructural features associated with oxidation
- Map microstructural features over time and temperature (isothermal kinetics)

Part II Examine the effect of environmental attack on fatigue resistance
- Pre-expose fatigue specimens using a subset of mapped conditions
- Examine effect of pre-exposure on fatigue life, crack initiation & propagation
- Correlate / model fatigue life to microstructural evolution

- Coupons & notched LCF specimens extracted from the rim of a fully heated forged disk, produced from HIP extruded powder billet
- Air Exposures: 704 °C – 815 °C up to 2,020 hours in a resistance furnace held isothermally, then air cooled
- NLCF testing at 704 °C, σ_{max}=855 Mpa, $\sigma_{min}/\sigma_{max}$= 0.05, 0.333 Hz with cylindrical notched specimens with K_t=2



Microstructural starting point



As-processed

Polycrystalline γ/γ' Ni-based superalloy

Large grain size for rim creep resistance
- <D>= 27 – 31 μm (ASTM 7.1-6.8)
- ALA= 105 – 125 μm (ASTM 3.5-3) (i.e. largest grain in cross section)

Produced by supersolvus heat treatment

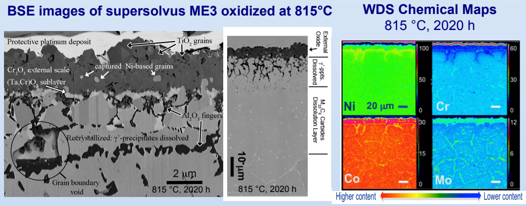
Minor phases for creep rupture
- Grain boundary (GB) strengtheners: Cr-rich $M_{23}C_6$ carbides (d <0.15 μm)
- Reside mostly in grains: MC carbides (0.15-0.7 μm) and M_3B_2 borides (0.4-1 μm)

γ'-precipitates for strength & fatigue resistance
- Tertiary γ', <d> = 18 – 39 nm
- Secondary γ', <d>= 190 –330 nm

$\gamma'(L1_2)$: $Ni_3(Al,Ti)$, strong Al partitioning
γ (fcc): Cr, Co, Mo partitioning

Microstructural features associated with oxidation

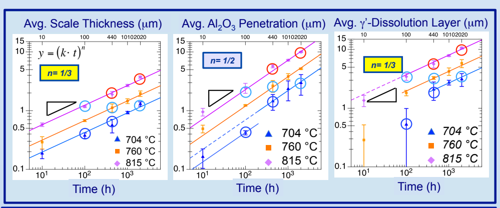
BSE images of supersolvus ME3 oxidized at 815°C — **WDS Chemical Maps** 815 °C, 2020 h



Like other high-Cr disk superalloys, a continuous Cr_2O_3 scale forms, with faceted, superficial TiO_2 grains at the exposed ME3 surface. Beneath the TiO_2-Cr_2O_3 scale, an internal subscale of branched Al_2O_3 extends into a layer where γ'- precipitates have been dissolved by Al depletion. This γ'-dissolution layer is recrystallized with finer grains and contains micron-sized voids at the grain boundaries. Throughout these layers and beyond, the $(Mo,Cr,Co)_{23}C_6$ carbides have been dissolved from the original grain boundaries via grain boundary diffusion of Cr that helps support scale growth.

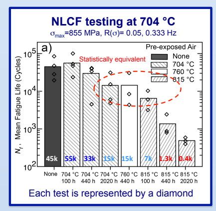
Select conditions for notched fatigue tests

Measure feature sizes over time & temperature space from cross sections



Slight exposures	Scale thickness < 0.6 μm:	704 °C for 100 h, 704 °C for 440 h
Moderate exposures	Invariant of temp.: Scale thickness ~1 μm & Al_2O_3 depths ~ 3 μm	815 °C for 100 h, 760 °C for 440 h, 704 °C for 2,020 h
Aggressive exposures	Scale thickness > 1.8 μm:	815 °C for 440 h, 815 °C for 2,020 h

Effect of prior exposures on fatigue response



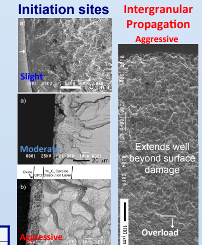
Each test is represented by a diamond

Fatigue response and failure modes

Slight	Moderate	Aggressive
Majority no debit, Bimodal dist.?	67%-85% Debit	97%-99% Debit
T initiation (single initiation)	Mixed I/T/O initiation (handful)	Oxide failure (multiple)
T propagation	Initial I and then T propagation	I propagation Extensive!!

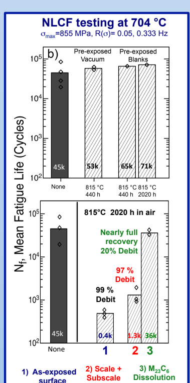
T= Transgranular, I= Intergranular, O= Oxide failure

Initiation sites — **Intergranular Propagation**

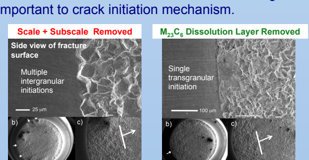


- Prior exposures in air significantly affect fatigue life.
- Debit is caused by damage at the surface not exposure temperature; as tests on moderate conditions produce equivalent lives.

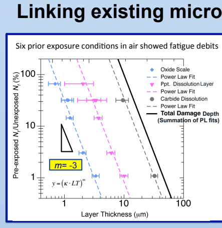
Additional fatigue testing



- Tests on specimens exposed in vacuum or exposed prior to machining showed no fatigue debit.
  - ➢Confirms debits observed for prior exposures in air are from environmental attack, not overaging during exposure.
- Tests on specimens, pre-exposed at 815°C 2020 h and where the oxide and subscale were removed and mechanically showed a marginal improvement (3X) in mean fatigue life compared to 815°C 2020h tests.
- When the $M_{23}C$ carbide dissolution layer was removed, a near full recovery was observed.
- The absence of carbides make GBs weak; layer removal tests establish that GB strength is important to crack initiation mechanism.

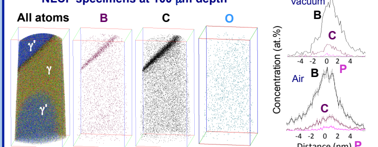


Linking existing microstructure to fatigue response



- With excellent fits to a simple power law, normalized fatigue life decays with thickness of oxide scale, γ'-dissolution layer, carbide dissolution layer, and, by inference, total damage depth with an exponent of -3.
- It follows that normalized fatigue life is proportional to $(time)^{-1}$ by substitution of cubic dependencies from cross sections.
- Removal experiments demonstrate that the fatigue response is governed by the total damage depth not individual layers.

Typical atom probe GB analysis from failed NLCF specimens at 100 μm depth



98 nm x 97 nm x 152 nm – Air exposure

Transgranular propagation
B=1.22 ± 0.20 at.%
C=0.28 ± 0.10 at.%
P=0.08 ± 0.04 at.%

Intergranular propagation
B=0.98+0.14 at.%
C=0.19 ± 0.06 at.%
P=0.059 ± 0.032 at.%

Aggressive exposures were found to degrade grain boundary strength well beyond the resulting surface damage. It is known that segregation of O or absence of B can can cause weak grain boundaries in similar nickel base superalloys [6], and therefore, it was hypothesized that long-range GB diffusion of these light species caused embrittlement. Precise measurement with atom probe tomography showed equivalent B, C, P and O chemistries for nondegraded and degraded grain boundaries, eliminating this possibility. Further work is planned.

Summary

- Static oxidation at potential service temperatures over extended periods was mapped for supersolvus ME3 from 704 °C to 815 °C.
- Cross-section evaluation uncovered complex near-surface damage, including extensive GB carbide dissolution.
- Fatigue debit reductions showed a power law correlation with total damage depth, that decay as $(TDD)^{-3}$ and by substitution, $(time)^{-1}$.
- Fatigue debit is independent of temperature and is caused by environmental surface damage from air exposure not overaging, while for specimens removal past carbide dissolution layer led to full recovery in fatigue life for aggressive prior exposures.
- For slight exposures, specimens failed from single surface cracks that initiated and propagated transgranularly, as exposures became more aggressive, multiple cracks initiated in surface oxide and propagated intergranularly for distances well beyond the total damage depth.

References: [1] JH Chen, PM Rogers, JA Little, Oxidation of Metals **47** (1997) 381. [2] A Encinas-Oropesa et al. in Superalloys 2008, 609. [3] TP Gabb et al. in Superalloys 2004, 269. [4] SD Antolovich, P Domas, JL Strudel, Met Trans A **10A** (1979) 1859. [5] Sudbrack et al. in Superalloys 2012, 863. [6] RC Reed, The Superalloys: Fundamentals and Applications (2006) 252.

Acknowledgements: The authors would like to acknowledge Dr. James Smialek and Dr. Mike Nathal (NASA-GRC) for valuable discussions, as well as Jeff Marshman (Carl Zeiss) for focused-ion beam assistance. Support of the NASA Aviation Safety Program, NASA Subsonic Fixed Wing program, NASA Undergraduate Student Research Program is also acknowledged. The atom probe work was performed at EMSL, a national scientific user facility at PNNL sponsored by the DOE's Office of Biological and Environmental Research, under a rapid access user proposal (#47633).