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Modelling of the fatigue crack growth of a coated single crystalline nickel-based superalloy under thermal mechanical loading

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ABSTRACT

The focus of this paper is the simulation of fatigue crack growth of the coated single crystalline nickel-based superalloy PWA 1484 under thermal mechanical loading. Thus, two physical models are superimposed in terms to firstly calculate the deformation behavior under instationary thermal and mechanical loading (TMF) and secondly to model crack propagation after initial brittle cracking of the coating layer on the basis of cyclic crack-tip opening displacement (CTOD). All material parameters implemented in the models were evaluated from monotonic isothermal tensile and creep tests as well as from isothermal low cycle fatigue (LCF) experiments. The calculated fatigue crack growth was validated by in situ crack growth measurements using the beachmark technique. Hence, crack propagation initiated by the brittle coating system closely to the experimental results using rectangular flat specimen geometry instead of corner-crack (CC) specimens. The comparison of the simulated lifetimes to the experimental results provides remarkable accuracy of the physically-based lifetime model.

1. Introduction

The aviation industry is under legal compulsion of current political circumstances like the Paris Agreement [1] to enhance efficiency. This is mainly realized by turbine improvements in terms of increasing the operating temperature in the combustion chamber. However, the increase in temperature is restricted by the high temperature capacity, i.e high temperature strength and oxidation resistance, of the turbine blade materials [2,3]. Today, single crystalline nickel-based superalloys are employed, because of their optimized creep resistance, which is attributed to the high volume fraction of hardening γ'-phase in the mechanically optimized solution hardened γ-matrix. Improved oxidation resistance is achieved by gas or pack aluminizing [4,5]. This aluminum enhancement leads to the formation of an intermetallic nickel-aluminide layer [6].

Under service conditions, turbine blades experience a superposition of complex thermal and mechanical loading [7]. Especially during start and shut-down of the engine very high temperature gradients occur and result in combination with the alternating mechanical loads in significant lifetime reductions. This loading situation is experimentally mimicked by thermo-mechanical fatigue (TMF) tests and is often regarded as the lifetime determining factor [8]. Under certain conditions, the presence of diffusion coatings can cause a reduced lifetime due to the brittleness of the intermetallic coating system at low temperatures [9,10]. In essence, the complex thermal and mechanical loading situation may lead to brittle cracking of the coating system, and thus, to drastical loss of fatigue life [11,12]. Given this scenario, a physically-based lifetime model for aluminide coated single crystalline nickel-base superalloys under thermo-mechanical loading was developed. An essential constraint is the brittle cracking of the coating system. Therefore, an approach for modeling of the deformation behavior by Wilhelm et al. [13] was used and superimposed by a modified crack propagation model based on the calculation of the cyclic crack-tip opening displacement as introduced by Schweizer et al. [14]. The simulation of the fatigue crack growth behavior was validated by in situ crack propagation measurements (beachmark technique) in the TMF tests.

2. Materials and methods

As substrate material a second-generation single crystalline nickel-based superalloy, PWA 1484, was selected. Specimens with a flat rectangular gage section were used in order to characterize the brittle crack initiation (Fig. 1). Brittle cracking of the coating system is shown in Fig. 2. All crystallographic orientations of the specimens were within

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Nomenclature

a	Crack length (m)
da/dN	Crack propagation per cycle (m/−−)
B_0	Material parameter from the Ramberg Osgood hardening law (1/MPa^n·s)
B_n	Material parameter from the Norton creep law (1/MPa^n·s)
c	Constant (m)
$C(t)$	$C(t)$-Integral (MPa/m)
CTOD	Crack-tip opening displacement for C(t)-Integral (m)
$d_{N'}$	Function of the hardening behavior from the HRR-field (−−)
E	Young's Modulus (GPa)
f_w	Width correction crack (−−)
$F_{C(t)}$	Function for the calculation of CTOD (m)
J	J-Integral (MPa/m)
K_I	Stress intensity factor (Mode I) (MPa $\sqrt{m}$)
K_{eff}	Effective stress intensity factor (Mode I) (MPa $\sqrt{m}$)
n	Exponent from the Norton creep law (−−)
N'	Exponent from the Ramberg Osgood hardening law (−−)
ε	Strain (−−)
σ	Stress (MPa)
σ_p	Threshold stress (MPa)
σ_y	Yield stress defined at 0.2% plastic strain (MPa)
BM	Beachmark
HRR	Refers to the singular crack-tip fields according to Hutchinson, Rice and Rosengren

4° off the ⟨0 0 1⟩ direction (Fig. 1: A-B axis). The solution heat treatment was performed according to Cetel and Duhl [15]. Subsequently, all specimens were coated (Fig. 3) in a chemical vapor deposition process. The main constituent of this diffusion coating is the intermetallic compound β- NiAl [16]. The Al concentration of the coating was 31 wt. %. The results of the coating thickness are listed in Table 1. The fracture strain as a function of temperature was determined in three-point bending tests by Affeldt [10], and a brittle-to-ductile transition temperature (DBTT) of 700 °C was determined.

To characterize the material response under the complex cyclic thermal and mechanical loading, TMF-tests were performed strictly corresponding to the TMF "Code of Practise (CoP)" [17]. For the TMF tests the temperature T was measured by a flattened, ribbon thermocouple (S type: Pt–PtRh10). The control thermocouple was attached centrically in the gauge length (not spolt-welded). In Fig. 4 the "out-of-phase (OP)" load case and the beachmark cycles (BM) used are shown schematically. For the OP TMF tests with a phase shift between the temperature und mechanical strain of 180°, the temperature range was 400–1100 °C with a heating and cooling rate of 10 K/s. The mechanical strain amplitude was set to 0.225%. The experimental program is listed in Table 2. All test parameters were kept constant, however, the stop criteria and the beachmark-to-TMF cycle rate N_{TMF}/N_{BM} (Table 2) and the number of load cycles per beachmark N_{BM} were defined on basis of an investigation from König et al. [18]. This approach allowed analysis of the microstructural evolution as well as the crack growth at different stages of fatigue.

For measurement of the fatigue crack propagation the beachmark method was employed. The OP-TMF cycle were interrupted after defined cycle intervals repeatedly, by keeping the minimum temperature at 400 °C and performing a beachmark sequence. For the beachmark cycles the maximum mechanical strain range was reduced to 0.3 of the given mechanical strain range whereas max. stress value was kept constant. The frequency was set to 2 Hz which is the limit of the high-temperature extensometer used. This technique changes the appearance of the fracture surface as a result of altering the loading and enables the measurement of fatigue crack growth through a post-test examination of the fracture surface [18]. Depending on the crack length, the number of the high-frequency cycles was adjusted to provide a clear optical appearance of the beachmarks while keeping the crack extension small. The location of beachmarks were measured using an optical microscope, and the crack growth rates da/dN were obtained from the spacings between the beachmarks.

The beachmark technique has the advantage, that there are no assumptions needed about the crack initiation and propagation stages. Moreover, this technique can be implemented easily in the test procedure without a change in the experimental set-up. The influence on the measured TMF-lives can be classified based on the width of the generated beachmarks. The maximum ratios of the crack length to the integrated beachmark width (all single beachmark widths' per test) constituted 3.5%. In addition, the beachmark cycles were performed at low temperature, where the overall deformation was primarily elastic and no creep effects occurred. Hence, the influence of the beachmark technique on the TMF-lives can be neglected, cf. Wei et al. [19]. Previous work on the alloy IN 718 and steel confirmed these assumptions [20].

3. Modelling

3.1. Deformation behavior

Wilhelm et al. [13] used a model introduced by Affeldt et al. [21] to calculate the thermo-mechanical deformation behavior of single

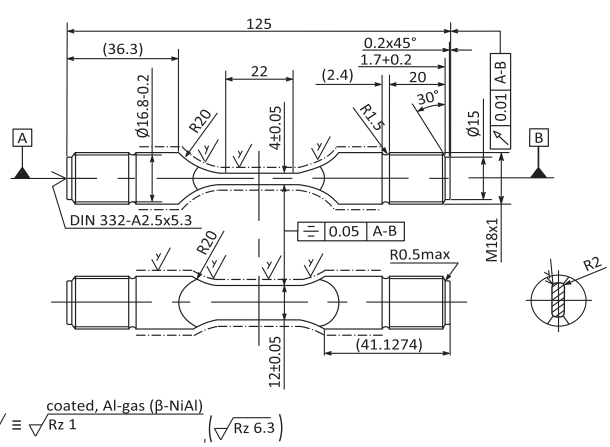

Fig. 1. Flat specimens according to TMF standard [39].

Fig. 2. Optical micrograph of brittle cracks in the coating system around the circumference of the specimen.