Analysis of Local Strain Evolution during Electron Beam Welding of Hot Crack Sensitive Nickel Base Conventionally Cast Alloy 247 LC CC

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1. Introduction

Nickel-based alloys are used in the field of aero and stationary gas turbines, where the increase in turbine exhaust temperature requires temperature-resistant materials to ensure a further reduction in nitrogen oxides. Superalloys of complex composition tend to form hot cracks when heat input by conventional heat treatment or welding processes is applied. The formation of precipitation-hardening phases by elements such as aluminum or titanium is relevant for various types of hot cracking such as liquation or solidification cracking. Electron beam welding is beneficial to reduce hot crack formation due to its variability of process control through beam deflection and definable heat input in a vacuum environment. The conventionally cast Alloy247 LC CC is not considered suitable for welding. Preliminary investigations with the electron beam at very low welding speeds (weld regimes outside conventionally used) show a reduction in hot cracking even with the very high heat input (Ref 1).

The approach from previous work on temperature field (Ref 2) and phase field simulation (Ref 3) was combined with the Rappaz’s method (Ref 4) to estimate the hot crack susceptibility in simple test samples welded by EBW with four different welding regimes. For this purpose, an implementation in a numerical simulation model using the following simplification has been realized:

\[
\frac{dk}{dT}_\text{crit} = \frac{\nabla T}{(1 + \beta)BT} \left( \frac{\lambda^2 \nabla T \Delta p_{\text{max}}}{180 \mu} - v_T \beta A \right)
\approx \left( \frac{\nabla T}{v_T} \right)_{\text{solid}} - k_2 \approx \frac{\lambda_{0.33}}{v_T} \approx \frac{1}{v_T^2}
\]

where the solidification isotherm velocity is expressed as ratio of temperature rate to thermal gradient: \( v_T = \frac{T}{\nabla T} \) and the empirical relation for the secondary dendrite arm ripening: \( \lambda_2 \sim \frac{T}{v_T^2} \). The results have shown a correlation between the regions with a solidification isotherm velocity above 40 mm s\(^{-1}\) and the locations of experimentally observed hot cracks, indeed, in the bottom region of the weld. Using increased welding speeds, a disagreement between simulation and experiment was observed at the seam shaft. Thus, an alternative approach is needed.

According to Prokhorov (Ref 5), the derivation of the hot cracking process cannot be represented exclusively by continuum mechanics formulations since the transition between solid and fluid is not continuous and not isotropic. Its defined technological strength references the deformation capacity within the crystallization process and considers the growing crystallization nuclei and their interstices. In the welding
process, the temperature boundaries for this growth are moving. The upper limit \( T_{\text{max}} \) within the brittle temperature range (BTR) is the temperature below liquidus where the crystal islands can first wedge together. The lower limit \( T_{\text{min}} \) below solidus forms the temperature of cohesivity at which also all liquid grain boundaries are solidified and able to absorb strains. Hot cracks are thus caused by the addition of internal deformations, which are either caused by design or can be generated in the test. The critical strain is reached when the total strain \( \varepsilon \) intersects the deformation function within the BTR, i.e., hollows from solidification shrinkages can no longer be filled with residual melt. For welding considerations, especially in beam applications such as laser or electron beam, the temperature rates and thus the strain rate is very high, and their determination requires complex application diagnostics. The temperature-related deformation rate can be determined according to

\[
\varepsilon_T = \frac{d\varepsilon}{dT} = \frac{d\varepsilon}{dt} \frac{dT}{dt}
\]

Prokhorov (Ref 5) thus specifies an engineering applicable hot crack critical range but does not consider the influences from solidification kinetics as well as morphology of the solidification front. Jüptner (Ref 6) uses an equivalent plastic deformation as intensity dimension of the deformation described by Prokhorov for a combined numerical–experimental evaluation of the hot cracking of thick-walled structural steels using laser-wire welds. The determined three-dimensional stress states correlate with the cracking phenomena, which represents the multiaxiality of the deformation capacity in this respect. Furthermore, the melt pool geometry and the joint shape (aspect ratio) influence the hot cracking (Ref 5, 6).

The aim of this work is to extend the understanding of weldability and hot crack sensibility of the turbine material Alloy 247 LC CC regarding the local thermo-mechanical conditions in the samples with the welding process parameters. For this purpose, a study using numerical thermo-mechanical welding simulation is considered. The focus is a correlation between the computational strain and the documented cracking phenomena in the different samples and thus to extend the method for constructive estimation of hot cracking tendency using CAE.

### 1.1 Test Procedure

A modified Houldcroft (Ref 7) specimen is used to compare hot cracks with strains in the welding temperature field. This allows a wide range of strain states to be simulated and quantifies the strain limits for the formations of hot cracks. The modification reverses the weld direction, as already shown by Shibahara (Ref 8), and adjusts the total design based on the material. The original Houldcroft test is based on an initial crack increasing during welding until the occurring weld stresses reach a minimum (weld direction from the area of stiffness to the one of flexibility). The length of the crack represents the hot crack sensitivity. For welding in direction to increased stiffness, hot cracking occurs upon reaching a defined stress level. Such space is designed by relieving gaps on the front and back side of the test sample (Figure 1). The tests are performed at welding speeds in the range between 0.5 and 10 mm s\(^{-1}\) and heat input between 797 and 117 J mm\(^{-1}\) followed by surface deformation measurement for validation of simulation and x-ray microtomography (\(\mu\text{CT}\)) analysis for documentation of hot cracking.

The samples have dimensions 100 mm of length, 60 mm of width and 6 mm of thickness. They were prepared by electrical discharge machining and finally examined for probable initial cracks and defects. The material is conventional cast Alloy 247 LC CC (Table 1). This alloy is classified as difficult to weld due to high amounts of eutectic phases formed during casting, which influence the width of the brittle temperature range.

The welding experiments were conducted with four welding regimes, Table 2, without applying any clamps or fixtures, which may additionally influence the stiffness.

### 1.2 Material Data

The material data for the simulation were created by assumed values and are shown in Figure 2. The assumptions are supported by measured data from similar alloys, theoretical models or according to single Gleeble tests. The specific heating and cooling conditions of the considered welding regimes are also taken into account.

For common nickel-based superalloys, important material data are published by Mills (Ref 9), e.g., the single-crystal solidified CMSX 4. Using the example of the wrought alloy IN718, Pottlacher (Ref 10) shows that the measurement methodology influences the common material data and suggests

| Table 1 | Nominal composition of Alloy247LC CC [wt.%] |
|---------|------------------------------------------|
| Co      | 9.2                                      |
| Mo      | 0.5                                      |
| W       | 9.5                                      |
| Ta      | 3.2                                      |
| Al      | 5.6                                      |
| Ti      | 0.7                                      |
| Cr      | 8.1                                      |
| Hf      | 1.4                                      |
| Ni      | Bal.                                     |

Figure 1  Adapted Houldcroft test sample (left), example for typical solidification cracks at \( v_s = 6 \text{ mm s}^{-1} \) and \( q_s = 160 \text{ J mm}^{-1} \) metallographically analyzed in longitudinal and cross section.
the use of the specific enthalpy. Torroba (Ref 11) illustrates enthalpy data for MAR-M247 (alloy similar to Alloy247 LC CC with additional carbide formers and reduced hafnium content). Michailov (Ref 12) shows an enthalpy range for comparable austenitic steels.

Considering the studied alloy and process conditions, dilatometer tests and hot tensile tests were carried out on the Gleeble3500 system, taking into account the thermal cycles from regime 1 and regime 4 (Table 2). In the dilatometer tests, cylindrical specimens (diameter 10 mm) are heated to 1300 °C and then cooled. In the hot tensile tests, “on-heating” and “on-cooling” cylindrical specimens with a diameter of the gauge section of 6 mm are used. Currently, limited data are obtained for the temperature interval between 1300 and 800 °C.

Based on the above information, the brittle temperature range (BTR) was assumed to be between 900 and 1370 °C, as well as liquidus at 1400 °C and solidus at 1340 °C, respectively. Accordingly, the assumed density, thermal conductivity, and specific heat capacity, used for temperature field simulation, are given as curves in Figure 2(a). Similarly, the thermal expansion coefficient, used for the structural analysis, is assumed as tabular function linear increasing from 11.01 K⁻¹ at room temperature to a limit value of 16.93 K⁻¹ at 1500 °C. In the used bilinear isotropic hardening model, the yield limit and the Young’s modulus are defined as curves (Figure 2(b)), and the tangent modulus is assumed to be 1 GPa below and 0.1 GPa above 1000 °C.

It is important to note that the strength properties of the base material are macroscopically less relevant than the grain structure and dendritic morphology. On the other hand, for the simulation, a continuum mechanical approach was applied. Thus, by designing the material model and interpreting the results, these both issues are specially considered. Finally, a series of test calculations with alternating material model parameter for the high temperatures were carried out to qualify the sensitivity of the simulation outcomes.

1.3 Temperature Field

The temperature field simulation was carried out by means of a transient finite element model, developed in a previous work (Ref 13). For the simulations, the ANSYS software was used. The finite element mesh was adjusted on the expected thermal and strain gradients and considering a symmetrical body. The material properties (thermal conductivity and volumetric heat capacity) were assumed as temperature dependent, as shown in Figure 2(a). An adapted heat source model (Ref 13) considers the bulge effect by a combination of a ring-shaped surface heat source (q₂) with a volume heat source along the beam direction (q₃), shown in Figure 3, and is used for the temperature field analysis.

On the exterior surfaces, a third type boundary condition considering radiation, i.e., satisfying the Stefan–Boltzmann law, was applied (Ref 13). The emission coefficient as well as the heat source model parameters was calibrated using measured temperature–time curves and micrographs, respectively.

The results analysis indicates that welding with lower welding speed causes a considerably wider temperature distribution, as compared exemplarily for regime 1 and regime 4 (see Table 2), i.e., welding speeds of 0.5 to 10 mm s⁻¹, in Figure 4, leading to overheating of the entire specimen body. The comparison of the weld pool shapes (assumed as 1370 °C isotherm) shows a clearly recognizable bulge on the pool rear side between a nail head region and a shaft region. A certain longitudinal extension on the weld pool rear side slightly below the center of the shaft could also be observed for all specimens.

| Parameter                  | Regime 1 (mm s⁻¹) | Regime 2 (mm s⁻¹) | Regime 3 (mm s⁻¹) | Regime 4 (mm s⁻¹) |
|----------------------------|------------------|------------------|------------------|------------------|
| Welding speed              | 0.5              | 2.0              | 6.0              | 10.0             |
| Heat input (J mm⁻³)        | 797              | 338              | 160              | 117              |

Figure 2 Material data assumed for the simulations: (a) volumetric heat capacity pc and thermal conductivity λ; (b) Young’s modulus, yield strength Rp0.2
In the domains free of macroscopic cracks, cooling rates up to 500 K s\(^{-1}\) can be observed. In the domains with macroscopic cracks larger than 1 mm, which particularly is the case in the shaft regions for tests with increased weld speeds, the cooling rates are above 2000 K s\(^{-1}\). This indicates corresponding high alternation of the thermal strain rate in the different samples.

### 1.4 Stress and Strain Field

The stress and strain field calculations were performed by steady-state structural analysis, considering small distortion, and weakly coupled with the transient thermal solution. An isotropic bilinear material model was used to describe the elastic–plastic behavior. The material properties (thermal expansion coefficient, Young’s modulus, yield stress and tangent modulus) are assumed as temperature dependent, see section Material Data and Figure 2a. The boundary conditions correspond to non-constrained symmetrical model.

To validate the results, the welding residual displacements fields were measured at the bottom and top of the samples and compared in the three main directions, and match over the entire area of visualization. The comparison of longitudinal, transverse and out-of-plane distortions, for example for regime 4 (see Table 2), is shown in Figure 5(a), (b) and (c), respectively, ref. also Figure 1.

The calculated fields of residual displacements match very well the measurement for all four tested welding regimes. Therefore, the numerical structural model was validated, and the analysis results of the single entities can be assumed as credible.

**Discussion of the Results**

The evaluation was based on the occurred stresses followed by the vectors of elastic–plastic strains. The first overall analysis of the results in the high-temperature domain indicates that the stresses and strains increase quickly at the beginning and then, along almost the entire welding path, do not alternate any more. The \(\mu\)CT examination of the welded samples also indicates a near equal crack susceptibility along the entire welding path. Therefore, only comparison between the different welding regimes, i.e., welding speeds, is essential.

The calculated stresses in the different samples correlate neither with the local position of the cracks nor in their orientation. Furthermore, the increase in tensile stresses with decreasing welding speed contradicts the trends found in the crack evaluation. Consequently, the cracks cannot be mathematically verified based on the stress field alone. One possible explanation is a temporal and spatial shift in the thermo-
mechanical processes, whereby in a reference volume, the surrounding material is still expanding, while the melt is already undergoing shrinkage.

In the analysis of plastic strains, starting from the ductility hypothesis of Prokhorov (Ref 5), the fields of the three vector components and their comparative strains are considered. Two further hypotheses are pursued. On the one hand, it is assumed that the cracks develop when a critical strain is reached in the respective directions, i.e., they are caused by the respective vector components. On the other hand, it is assumed that the plastic capacity is reached when a critical strain is exceeded, regardless of the sign and direction (equivalent strain). In longitudinal direction, according to Michailov (Ref 14), plastic compression occurs in the weld area during heating. At the beginning of cooling, the thermally induced contraction causes a stress reduction without changing the already accumulated plastic compression. With further cooling/contraction, tensile stress increases until the yield point is reached, and a plastic strain reversal starts. In the case of the weld specimen considered, this “second” cooling phase starts below the BTR and thus does not include any change in the longitudinal plastic strain. The transverse component of the plastic strain develops identically during heating (locally and temporarily). A plastic compression develops in the transverse direction, which, however, remains approximately unchanged after the peak temperature is reached. An exception is at the seam interface, but there are no cracks in this region. Thus, no relevant plastic expansion in the transverse direction is found in the BTR. In the z-direction (along the beam, see also coordinate system on Figure 1), expansion occurs during heating and the heated metal contracts. Immediately at the cooling, the contraction in the z-direction is influenced by the metal and presses in the other directions. Consequently, contraction in the nearly stress-free z-direction is controlled by the other two strain components, causing an accumulation of plastic expansion. In this process, the complex geometry of the temperature field has a key role. The simulation results show that this aspect is clearly pronounced and leads to a continuous increase in the plastic strain maxima in the local areas with the progressive cooling along the seam. In the high-temperature range, the strength of the material is low; therefore, almost all thermal strain passes into the plastic. Pure elastic strains start significantly below the ductility recovery temperature (DRT) during cooling and proceed at extremely low levels. This indicates the ductility hypothesis postulated by Prokhorov (Ref 5) corresponding to the ductility hypothesis and not fails due to tensile stress. For this reason, the plastic strain distribution was considered in the analysis. The equivalent strain $\varepsilon_{eq}^{pl}$ (von Misses) and the strain in the z-direction $\varepsilon_{z}^{pl}$ are relevant. It should be emphasized that over the entire BTR and for all welding regimes, the z-component and the first principal component of the plastic strain vector was found to be coincident. A clear correlation with the evaluation of the crack phenomena by means of plastic strain magnitude cannot be given. However, the analyses indicate that micro-cracks correlate with the magnitude of equivalent plastic strain and macro-cracks with the z-component of plastic strain. The distribution of strains in the z-direction within the BTR is shown in Figure 6 exemplary for regime 3, i.e., $v_{c}=6$ mm s$^{-1}$, (Table 2).

To evaluate magnitudes for hot cracking of the electron beam welds, relevant strain and temperature progress was compared with the positions of the crack locations. A qualitative comparison for the hot crack occurrence in the welded samples and the strain change in the thermal cycle is achieved by the relevant strain components vector, elastic or plastic. The analysis is performed by comparison of the strain and thermal trends during cooling, see also Figure 7. A special aspect for this assessment is the utilization of the temperature $(T_{C})$ in the calculation as the first occurrence of strain, i.e., strain recovery temperature, taken over from Herold and Streitenberger (Ref 15). In the current work, this reference temperature was assumed to be the calculated 1400°C isotherm, as it supports optimal the comparison of the results from all four tested welding regimes.

The hot cracking hypothesis can be represented by three characteristic profiles of z-strain component in the temperature range of the BTR (curves I to III), Figure 8. The value of the critical strain $\varepsilon_{C}^{pl}$ can be determined if the volume passes through the melting temperature and BTR without hot cracking during cooling. The temperature-related deformation rate $\varepsilon_{T,cr}$

![Figure 5](image-url) Field of residual displacements exemplary top side for regime 4, comparison of simulation (always left) and measurement results (right): (a) x-displacement $U_x$, (b) y-displacement $U_y$, (c) z-displacement $U_z$. 
starting at the reference temperature $T_U$ is a material model-specific limit. When both criteria are exceeded during cooling, a hot crack formation is expected, Figure 8 (II). Exceeding only one criterion is not enough for a hot crack formation, Figure 8 (I) and (III). This study assumes a constant gradient $\varepsilon_T$, while Prokhorov (Ref 5) indirectly assumes different rates at different welding speeds based on the melt pool form. For a direct comparison, these melt pool shapes should be identical, which is not possible in beam welding due to the aspect ratio and the presence of the nail head. Jüptner (Ref 6) describes this tendency as thermodeformation (temperature-related deformation rate $\partial \varepsilon / \partial T$) and confirms the influence of the weld shape in this regard.

The evaluation area in the cross section is separated into different levels of weld depth (A...D) and split according to the center of the melt volume and near the fusion line (1,2), as shown in Figure 9. These areas were compared with the analysis of the statistical occurrence of hot cracks, see Table 3, and the plastic strain curves were generated at significant points of interest. In a further step, the assumed reference temperature $T_U$ was applied to the calculated strains (denoted further as comparative strain) in order to determine the strain behavior criterion.
The analysis of critical strain states is based on the simulation. Since the results are based on a comparison with a real hot cracking behavior, additional material-related features (different forms of hot cracking in nickel-based superalloys) have been included and categorized as best as possible. Figure 10 shows the comparative plastic strain in z-direction in the temperature range of the BTR for all tested welding regimes and critical strain curves as maximum values. In addition, the smallest comparative strain is shown for regime 4 (Table 2), i.e., $v_s = 10 \text{mm s}^{-1}$. Using the compared crack statistics, the critical value for the plastic strain is determined. For the regime 1, i.e., $v_s = 0.5 \text{mm s}^{-1}$ no cracks were observed; therefore, a robust strain behavior can be assumed.

The first criterion for the formation of hot cracks is the critical plastic strain $\varepsilon_{kr}^{pl}$ in the z-direction (direction of the electron beam) after recovery of the positive strain during cooling. For electron beam welded butt joints of conventionally cast Alloy 247 LC CC, i.e., created material data base, a value of $\varepsilon_k^{pl}$ of 0.021 was found. This ensures an almost crack-free weld at regime 1, i.e., $v_s = 0.5 \text{mm s}^{-1}$.

The second criterion refers to the increase in the temperature-related deformation rate $\dot{\varepsilon}_{kr} = \frac{d\varepsilon}{dT}$ in combination with the plastic strain after leaving the BTR. In this case, a very sharp increase above the limit $\varepsilon_{kr}$ does not lead to cracking at final strain below $\varepsilon_k^{pl}$. The combination of both values, $\dot{\varepsilon}_{kr} > \varepsilon_{kr}$ and $\varepsilon_k^{pl} > \varepsilon_k^{pl}$, leads to hot cracking due to exhaustion of the material. In this case, the considered critical temperature-related deformation rate is found to be $\dot{\varepsilon}_{kr}$ at $-1.67 \times 10^{-4} \text{K}^{-1}$. For illustration, the calculated plastic strains were referenced to the critical strains, $\Delta \varepsilon_0 = \varepsilon_0 - \varepsilon_{kr}$ and $\Delta \varepsilon_k = \varepsilon_k^{pl} - \varepsilon_k^{pl}$, see Figure 11.

A more sustainable characterization can be obtained by consideration of an additional, third criterion, which is enclosed in the thermal part of the simulation model. For the current application, it refers to the results reported in the previous work (Ref 2). In that case, the third criterion limits the strain related to the isothermal velocity of dendritic solidification to a maximum of $v_T < 40 \text{mm s}^{-1}$. The application of these three criteria on all four tested regimes (ref. Table 2) is shown in Figure 9. Clear correlations can be observed between the documented cracks (black points) and the individual criteria. In the domain of the nail head more sensible is the third criterion (white points), where the other two, i.e., plastic strain magnitude and strain rate, do not recognize the cracks. In this area, the cracks are classified as micro-cracks, i.e., small than 1-mm length. In contrast, in the domains of shaft, more sensible is the first criterion—plastic strain magnitude (black line). For the transition between shaft and heat, a best correlation gives the third criterion (gray line). In the shaft and in the transition domains, the cracks are classified as macro-cracks, i.e., cracks with length of several millimeters.

The applied continuum mechanical calculation indicates the location where the critical conditions for a crack initiation are present. In the evaluation of experimental results, the location of the cracks is averaged along their surface. This, especially for the macroscopic cracks with length of several millimeters can produce some deviation.

In addition, by the documentation of the observed cracks, so far, a $\mu$CT was used, a clear separation between solidification, liquidation, and ductility dip cracks could not be done. Thus, in the analysis, only separation based on the crack length was taken into account.

### 2. Conclusions

Developed for increased castability, the high-temperature nickel-based material Alloy247 LC CC remains challenging to weld due to its high susceptibility to hot cracking. Conventional process adaptations to reduce hot cracking must be extended for electron beam welding to take the special weld shape of the melt pool into account. In this study, a continuum mechanical approach based on thermo-mechanical welding simulation was used to analyze the stress–strain development in the BTR and thus to clarify the relations between welding process parameters, material, and hot crack susceptibility. Special attention was paid to the creation of a consistent material model and the influence of the individual material parameters on the computational outcomes.

Four welding regimes with different welding speed, i.e., heat input, were used to produce samples with different level of cracking. The analysis of the stress and strain vectors in every sample focuses on the components sign (expansion or compression), their magnitude as well as on the intensity of the change versus the temperature rate. The overall analysis of the results indicates that in the current cases, the plastic strains in direction toward the thickness and the first principal plastic strain coincide and thus can be equally used to explain the observed cracking phenomena. The second principal plastic strain component

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**Table 3 Statistical analysis of the probability for hot crack formation at different evaluation levels and weld speeds**

| $V_{s}$ (mm s$^{-1}$) | A1 | A2 | B1 | B2 | C1 | C2 | D1 | D2 |
|-----------------------|----|----|----|----|----|----|----|----|
| 0.5                   | -  | -  | -  | -  | -  | -  | -  | -  |
| 2.0                   | +  | +  | +  | +  | +  | -  | +  | +  |
| 6.0                   | +  | +  | +  | +  | +  | -  | +  | +  |
| 10.0                  | +  | +  | +  | +  | +  | +  | +  | +  |

Probability: – very low – low + few ++ high
coincides with transverse strain, but its magnitude is considerably smaller and in some domains of BTR even compressive. An elastic tension in the BTR could not be observed.

Prokhorov’s theory for the design of technological strength from minimum strain and deformation rate was confirmed but needs to be extended and combined with the resulting localized plastic strain and strain rate as criterion. The deformation rate can exceed the critical values without producing a crack if the strain remains below the critical level.

However, in some local areas, especially in the nail head region where the cracks are small and dispersed, a discrepancy between the evolution of the plastic strain (equivalent or z-component) and the observed cracks was detected, where an evaluation based on the strain expressed through the solidification isotherm velocity appears more appropriate.

By superimposing the plastic strain magnitude and rate from the mechanical approach and the solidification isotherm velocity from the thermal analysis, a correlation between both micro- and macro-cracks for entire weld cross section could be obtained.

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