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# **ISOTHERMAL LOW CYCLE FATIGUE OF A LOST FOAM CAST AL-SI-CU ALLOY: STUDY OF THE DAMAGE MECHANISMS WITH SYNCHROTRON X-RAY TOMOGRAPHY AND DIGITAL VOLUME CORRELATION**

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## **ABSTRACT**

Al-Si-Cu alloys manufactured by the Lost Foam Casting process have a complex microstructure. In order to study its role on damage under loading conditions representative of the in-service conditions of automotive cylinder heads, in situ Low Cycle Fatigue tests monitored by synchrotron X-ray tomography have been performed at room and high temperature (250°C). The initiation and the first propagation stages of short fatigue cracks have been observed in the interior of the samples. Those observations were correlated with the 3D strain fields obtained from Digital Volume Correlation. Cracks nucleate usually at the first loading cycle at the sharp edges of subsurface pores due to their notch effect that induces a strain localization but also at hard inclusions located in the pores neighbourhood. Then, crack propagation proceeds through hard inclusions where the von Mises cumulated strain increases. This scenario is however temperature dependent. At room temperature, crack initiation is porosity driven while propagation is correlated with the presence of hard phases in the interdendritic space. At high temperature, massive fracture of eutectic Silicon was observed. Crack propagation is then dictated by the coalescence of these microcracks in the most strained area of the specimen, i.e. between or close to large pores.

## **KEYWORDS**

Cast Al Si alloys, X-ray tomography, Digital Volume Correlation, in-situ fatigue tests, damage mechanisms, high temperature Low Cycle Fatigue

## **INTRODUCTION**

In order to meet environmental objectives, the automotive industry has developed an engine downsizing strategy to reduce fuel consumption. The use of a Lost Foam Casting (LFC) process allows casting cylinder heads to their near final dimensions hence at a reduced cost and with a minimized waste of material. However, the LFC process also implies a low cooling rate and the use of an evaporative polystyrene pattern. While the former induces a coarse

microstructure in Aluminium Silicon alloy, characterized with a Secondary Dendrite Arm Spacing (SDAS) of about 75 $\mu$ m, the latter induces a higher porosity content than conventional gravity Die Casting (DC) process. As a result, the fatigue life of an A319 alloy can be strongly reduced when comparing the LFC alloy to the DC alloy although the cyclic properties of the aluminium matrix remain the same, i.e. the hysteresis stress-strain loops for stabilized cycles are the same [1].

The objective of this work is to study the role of the complex microstructure inherited from the LFC process on fatigue damage in isothermal Low Cycle Fatigue (LCF) at two temperatures that are representative of the minimum and maximum temperature experienced by the cylinder heads in service conditions: room temperature and 250°C.

A dedicated experimental protocol was set up to allow in-situ monitoring of fatigue crack nucleation and growth in the bulk of samples using synchrotron X-ray tomography and a miniature fatigue test rig described in [2]. The tomography images that were acquired at the minimum load of the first cycle and at the maximum load of subsequent cycles were then analysed using Digital Volume Correlation (DVC) to measure the three-dimensional von Mises strain fields. The evolution of these strain fields during cycling helps to understand the damage mechanisms observed.

## SPECIMEN, MATERIAL AND TESTING

### Material

The studied alloy is an Aluminium Silicon Copper (close to A319 alloy) alloy having the composition shown in Table 1. Fatigue specimens were cut out of the most critical areas of LFC cylinder heads, i.e. the fire deck.

Al	Si	Fe	Cu	Mn	Mg	Ni	Zn	Pb	Sn	Ti	Cr	Sr
bal.	7.85-7.90	0.30-0.31	3.05	0.19	0.28-0.29	0.018	0.16	<0.015	<0.015	0.097-0.098	0.015-0.016	~150ppm

**Table 1:** Chemical composition range of A319 alloy in this study (in weight %)

The microstructure of A319 alloy consists, on the one hand, in large aluminium dendrites (SDAS  $\approx$  75 $\mu$ m), and large microshrinkage pores, and on the other hand, in hard inclusions, i.e. eutectic Si, eutectic Al<sub>2</sub>Cu and iron-based intermetallic particles, within the interdendritic space. The pore population mainly consists in numerous small spherical pores with an average Feret diameter of about 30 $\mu$ m and a few large and tortuous microshrinkage cavities with a Feret diameter that can be as large as 1.2mm; see reference [3] for a more complete description. As a secondary alloy the A319 composition may include up to 0.6w% of Fe that forms iron based intermetallics upon cooling. The negative influence of these intermetallics has been largely discussed in the literature but their influence is still complex to predict as they are reported to facilitate crack initiation but also to hinder crack propagation [4]. Two kinds of iron intermetallic are observed in the present alloy: the so-called beta phase Al<sub>5</sub>FeSi with its platelet shape as well as the alpha phase Al(Fe,Mn)Si with its Chinese script morphology. Both form 3D interconnected clusters together with the eutectic Al<sub>2</sub>Cu [2]. However the eutectic Si, which shows a coarse and acicular morphology, forms the largest networks observed.

## Testing

In order to study the influence of the LFC microstructure on fatigue damage, in situ LCF tests monitored by synchrotron X-ray tomography have been performed at room and high temperature (250°C).

To achieve a voxel size relevant for the size of the microstructure and for imaging cracks, small dog-bone specimens with a millimetric square cross section were used. The parameters for the two fatigue campaigns are summarized in Table 2. Because the size of the samples is comparable to that of the largest pores, a selection of specimens based upon preliminary laboratory X-ray tomography characterization of pores in the sample was done as explained in [3].

	Fatigue parameters				X-ray tomography parameters				DVC parameters	
	Specimen cross section (mm×mm)	Max. applied strain (%)	Max. applied stress (MPa)	Stress ratio	Temperature	Voxel size (μm)	X-ray beam energy (keV)	Scan duration	Spatial resolution (μm)	Uncertainty (μm)
SLS	2.6×2.6	0.14	130	~0.1	20°C	1.625	25	7min	13	0.112
ESRF		0.4	70		250°C	2.75	35	45s	11	0.160

Table 2: Parameters used for the in-situ fatigue tests and their analysis with DVC

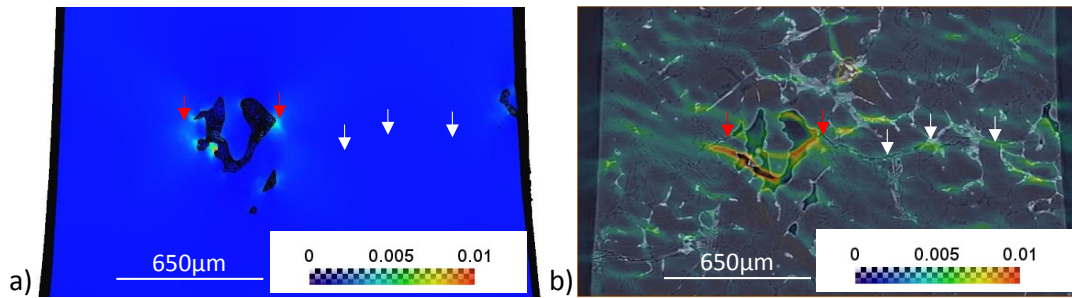
Fatigue tests were conducted using the fatigue test rig described in [2] first at room temperature at the Swiss Light Synchrotron (TOMCAT beamline) then, at 250°C at the European Synchrotron Radiation Facility (ID19 beamline) using a fixed lamp furnace that allows heating the specimen without masking the X-ray beam. At both temperatures, tomography acquisition was performed with the specimen held under load (in displacement control) and after relaxation has occurred. The loading conditions and parameters for X-ray tomography acquisition are summarized in Table 2. X-ray tomography was performed in the so-called edge enhancement mode that serves two goals: first it allows eutectic Si to be distinguished from the Al matrix although their X-ray absorption coefficients are almost the same, secondly it allows a better detection of cracks. Several specimens were tested at each temperature but only two specimens, one for each temperature, are presented herein. They are representative of the mechanisms observed in the rest of the samples.

The microstructure of the LFC Al-Si-Cu alloy as observed by synchrotron X-ray tomography forms a rich natural speckle, suitable to perform DVC at a spatial resolution that is high as compared to the size of the microstructure. Using an in-house software, YaDICs [5], a spatial resolution of the measured displacement field as small as about 12μm could be obtained, i.e. the displacement was measured at the nodes of a regular 3D mesh with cubic elements having an edge of 12μm. See Ref. [6] for a detailed explanation of DVC computation and precision assessment. Due to the difference in voxel size at the two synchrotron sources, the DVC spatial resolution was 8 voxels (13μm) with an uncertainty of 0.069 voxels (0.112μm) and 4 voxels (11μm) with an uncertainty of 0.058 voxels (0.16μm) for SLS and ESRF images respectively.

## EXPERIMENTAL RESULTS

The initiation and the first propagation stages of short ( $\sim 100 \mu\text{m}$ ) fatigue cracks have been observed in the interior of the samples. Those observations can be correlated with the 3D strain fields obtained from DVC.

Cracks nucleate usually at the very first loading cycle at the sharp edges of a subsurface pore as shown in Fig. 1. The notch effect at the corners of a microshrinkage cavity induces a strain localization that can be computed using Finite Element Modelling (FEM) as shown in Fig. 1a. FEM was performed using a 3D mesh of the porous matrix as obtained from X-ray tomography and assuming a homogeneous elastoplastic behaviour of the matrix with isotropic hardening (see ref. [3] for more details); hard inclusions are not taken into account. The largest computed strains in monotonic loading correlate well with the crack initiation sites observed after 2 cycles (see the red arrows in Fig. 1a). Strain localization at the pores is also visible in the strain field measured by DVC (Fig. 1b). However, apart from the pores, strain localization is also observed at hard inclusions at locations that coincide with the crack path observed just before failure of the specimen at 10500 cycles (see the white arrows Fig. 1b).

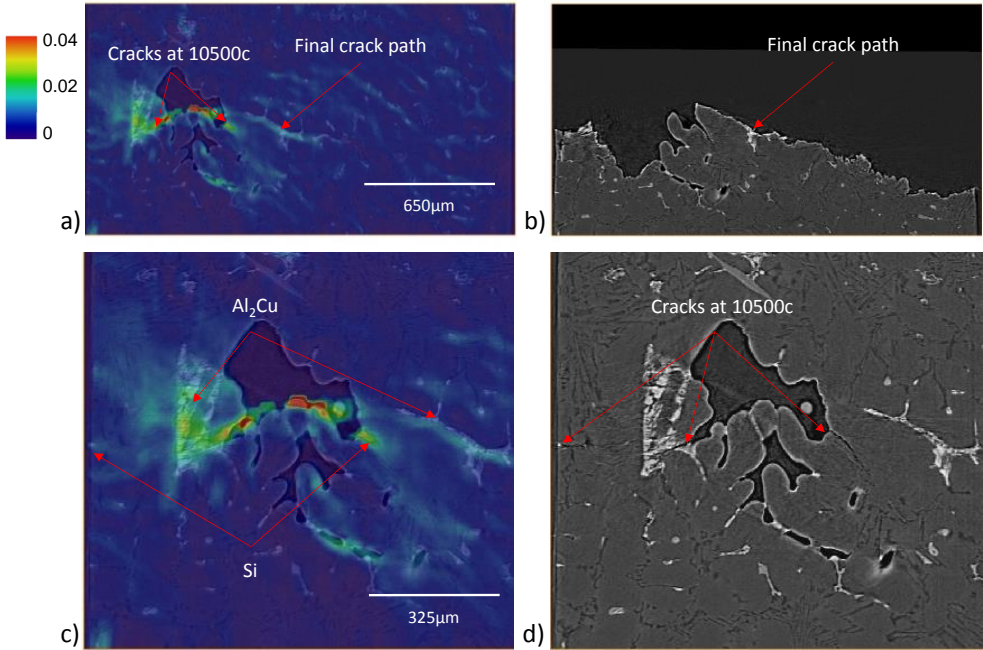


**Fig. 1:** Crack initiation at a subsurface pore about  $400 \mu\text{m}$  below the surface at room temperature: (a) FE equivalent plastic strain after 3 cycles and (b) DVC cumulated von Mises strain measured after 2 cycles; Red (resp. white) arrows indicate cracks initiated after 2 cycles (resp. 10500 cycles) and microstructure of the cracked specimen at 10500c is overlaid.

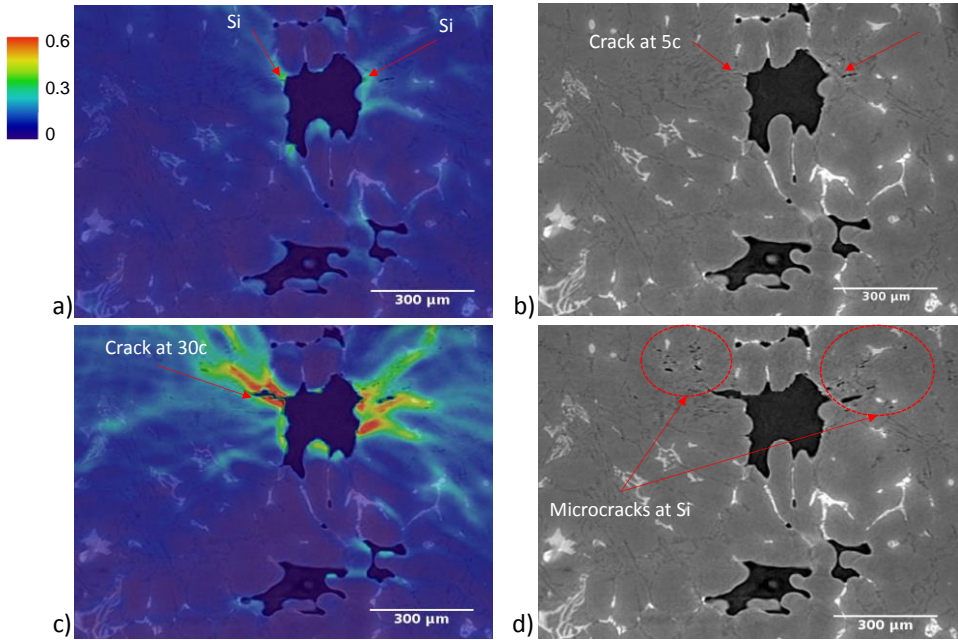
Crack propagation is thus observed to proceed through hard inclusions where the von Mises cumulated strain measured with DVC progressively increases until failure. Fig. 2 shows the same pore as in Fig. 1 seen from a different orientation. After 1000 cycles, no cracks can be detected in this slice (Fig. 2a and c). However, larger strains are observed even away from the pores edges, especially on the right hand side at a eutectic Si particle (see arrows) but also on large clusters of eutectic  $\text{Al}_2\text{Cu}$  (see arrows) on the left and on the right hand side of the pore. These strain localizations are consistent with the cracks at the following cycles (Fig. 2d) and with the final crack path (Fig. 2b).

At  $250^\circ\text{C}$ , cracks still initiate at the first loading cycle at the sharp edges of pores where the measured strain localization is a result of the pore shape (Fig. 3a). However, massive fracture of eutectic Silicon is also observed away from the main crack initiation area and in the neighbourhood of the pore as shown in Fig. 3b. These microcracks grow/open with further cyclic loading (Fig. 3c) and eventually coalesce. The large “strain localization” in Fig. 3c after 30 cycles reveals the shape of this coalesced cracks; the level of strain is no more realistic as the presence of microcracks (Fig. 3d) induces discontinuities in the displacement field. Crack propagation is then dictated by the coalescence of these microcracks at eutectic Si in the most strained area of the specimen, i.e. between or close to large pores. Strain fields are still under

analysis to understand the reason why eutectic Si behaves differently between room and high temperature tests.



**Fig. 2:** Crack propagation at room temperature in a slice about 800 μm below the surface: (a) the DVC cumulated von Mises strain after 1000 cycles (with microstructure overlaid) correlates well with the crack path shown in (b) at 10500 cycles; failure occurred just after 10500 cycles in this specimen. A focus on the pore where cracks nucleated (c) shows that, although no cracks is visible in the overlaid microstructure, strain localization in hard inclusions correspond to the cracks observed after 10500 cycles as shown in (d).



**Fig. 3:** Crack propagation at 250°C in a slice about 900 μm below the surface: (a) the DVC cumulated von Mises strain after 5 cycles (with microstructure overlaid) shows strain localization at the pore corners (arrows) where (b) cracks initiated at eutectic Si. After 30 cycles, a large strain localization in the cumulated strain (c) correlates well with the final crack path while, in the strain localization area, several microcracks are detected in Si in (d).

## CONCLUSIONS

In the studied LFC A319 alloy, during LCF tests, cracks are observed to nucleate at the first loading cycle at the sharp edges of subsurface pores due to their notch effect that induces a strain localization. Then, crack propagation proceeds through hard inclusions where the von Mises cumulated strain measured with DVC is observed to increase as the number of fatigue cycles is increased. This general scenario is however temperature dependent. At room temperature, crack initiation is porosity driven while propagation is correlated with the presence of hard phases in the interdendritic space. On the contrary, at high temperature, massive fracture of eutectic Silicon was observed and cracks propagation corresponds to the coalescence of these microcracks in the most strained area of the specimen, i.e. between or close to large pores.

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