



Martinez Ubeda, A. I., Warren, X., Griffiths, I., & Flewitt, P. E. J. (2016). The role of prior fabrication and in service thermal ageing on the creep life of AISI Type 316 stainless steel components. In J. Toribio, V. Mantič, A. Sáez, & M. H. F. Aliabadi (Eds.), *Advances in Fracture and Damage Mechanics XV: Selected, peer reviewed papers from the 15th International Conference on Fracture and Damage Mechanics, September 14-16, 2016, Alicante, Spain* (pp. 1-4). (Key Engineering Materials; Vol. 713). Trans Tech Publications Inc. <https://doi.org/10.4028/www.scientific.net/KEM.713.1>

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The role of prior fabrication and in service thermal ageing on the creep life of AISI Type 316 stainless steel components

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Keywords: Creep, Austenitic stainless steel, phosphorus segregation, Chi phase, M₂₃C₆ carbide.

Abstract. A significant factor in creep life of AISI Type 316H austenitic stainless steel components such as headers, and tubes is the initial microstructure. These components typically have a comparable specified composition but different thermo-mechanical fabrication histories. The variations in composition within the nominal range result in initial microstructures which become increasingly divergent during ageing. In this paper we explore effect of these contributions on the long term service aged microstructure and discuss the resulting impacts on the overall creep life of these components. The microstructure of specific regions has been characterised with a range of techniques, including high resolution transmission electron microscopy imaging and chemical analyses undertaken using a JEOL ARM instrument operating at 200 KeV fitted with an energy dispersive spectrometer. This provides a unique identification of the service aged precipitates and the distribution of alloying and impurity elements. The results are discussed with respect to the initiation of creep cavities and the associated creep damage accumulation in the context of lifetime assessment of these AISI Type 316H austenitic stainless steel boiler components.

Introduction

The UK's Advanced Gas Cooled Reactor (AGR) nuclear power generating plant operates at temperatures up to 550°C, where creep life is important. Creep deformation and fracture are controlled by a series of interacting factors, which need to be considered to evaluate service life [1]. In the context of AGR boiler components this specifically relates to the reduction of integrity arising from time and temperature dependent microstructural parameters and the resulting impact on creep strength and creep damage. Although the casts of steel for example AISI Type 316H austenite stainless steel used to manufacture different plant components may have similar specific compositions, fabrication processes vary. Large components such as boiler headers typically have very little subsequent fabrication following casting, giving an enhanced likelihood of preserving the original dendritic structure compared to highly worked material such as piping. The microstructure of austenitic stainless steels has been previously observed to evolve significantly during thermal ageing [1], with secondary phase precipitates nucleating throughout the microstructure and growing over time. The phases which form are dependent on both the composition and fabrication history of the component as well as the period of thermal exposure [2]. Warren *et al* have shown that ferrite precipitates in an ex-service steel were associated with creep cavities [3], and thus that phase evolution can enhance creep damage accumulation. In this paper we explore the role of prior fabrication further.

Experimental

The two components studied (a boiler header and a bifurcation tube) were both AISI Type 316H austenitic stainless steel components from advanced gas cooled reactors (AGR), provided by EDF Energy Ltd. During service the boiler header specimen (60 mm thick) experienced 65,015 hours operation in the temperature range 490°C to 530°C, followed by a post-service laboratory heat treatment of 22,100 hours at 550°C. The bifurcation was fabricated by bending a length of tube (3 mm thick) through an angle of 150° and cutting the extrados of the bend to leave an oval opening. The bifurcation was in service for approximately 160,000 hours at about 505 °C. The main elemental composition of the header and the bifurcation tube are given in Table 1. The extrusion process was higher in the case of the bifurcation tubes than for the header material.

Due to the differing thermo-mechanical fabrication routes, the microstructure of the thick walled header and the thin walled bifurcation tube varies significantly; with the header showing discrete Cr and Mo enriched regions originating from insufficient homogenisation of the original cast microstructure, whilst the parent material of the tube is significantly more homogenous.

Table 1: Composition of the boiler header and the bifurcation tube [wt%]												
Sample	C	Si	Mn	P	S	Cr	Mo	Ni	B	Co	Cu	Fe
Ex-service boiler header	0.06	0.40	1.98	0.02	0.01	17.17	2.19	11.83	0.01	0.10	-	66.23
Ex-service bifurcation tube	0.05	0.41	1.47	0.02	0.01	16.60	2.43	11.70	-	0.05	0.11	balance

Table 1: Composition of the ex-service Type 316H austenitic stainless steel header and the ex-service Type 316 austenitic stainless steel tubing [wt%].

Transmission electron microscopy (TEM) and scanning transmission electron microscopy (STEM) were performed using a JEOL-ARM 200F, operating at 200 kV and fitted with an energy dispersive X-ray spectroscopy (STEM-EDX) with a 100 mm² Centurion detector and NSS version 3.2 analysis software.

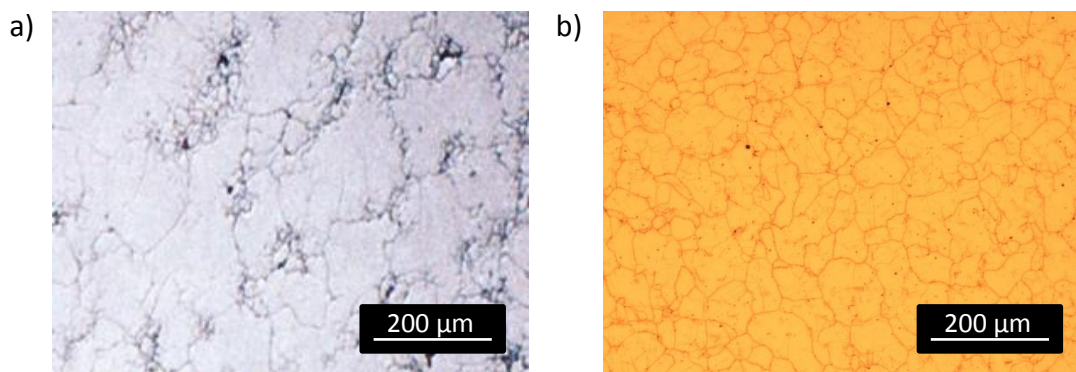


Figure 1: Optical micrograph showing a) the distribution of fine grained austenite/ferrite/ $M_{23}C_6$ carbide in the Cr and Mo rich residual casting regions in the ex-service Type 316H boiler header [4] and b) the homogenous parent grain distribution of the bifurcation tube

Results

The Cr and Mo enriched regions on the boiler header, Figure 1a, were typically widely distributed and had a different range of secondary phases compared with the bulk material [3]. Following TEM analysis, these regions of the boiler header were found to contain a complex mix of secondary phases including ferrite, chi phase, G-phase, γ' , and $M_{23}C_6$ carbide [3,5]. EDX microanalysis showed that the only significant elemental segregation occurred at chi phase-austenite interphase boundary, Figure 2a, where phosphorus was enriched by 0.7 wt%, Figure 2c. Phosphorus segregation at austenite-ferrite boundaries has been predicted [5], however was not observed. The larger lattice mismatch between austenite-chi phase and austenite-ferrite unit cells (chi phase $a = 0.795\text{-}0.845$ nm [6] vs austenite $a = 0.358$ nm [7] and ferrite $a = 0.285\text{-}0.289$ nm) would be expected to favour segregation kinetics to the chi-austenite interphase.

The bifurcation tubes showed $M_{23}C_6$ carbide, ferrite and G-phase after 160,000 hours ageing, which was similar to the bulk regions of the boiler header, Figure 2b. In this case, EDX microanalysis showed segregation occurred at $M_{23}C_6$ carbide-austenite interphase boundary with a typical value of 12 wt%, Figure 2d.

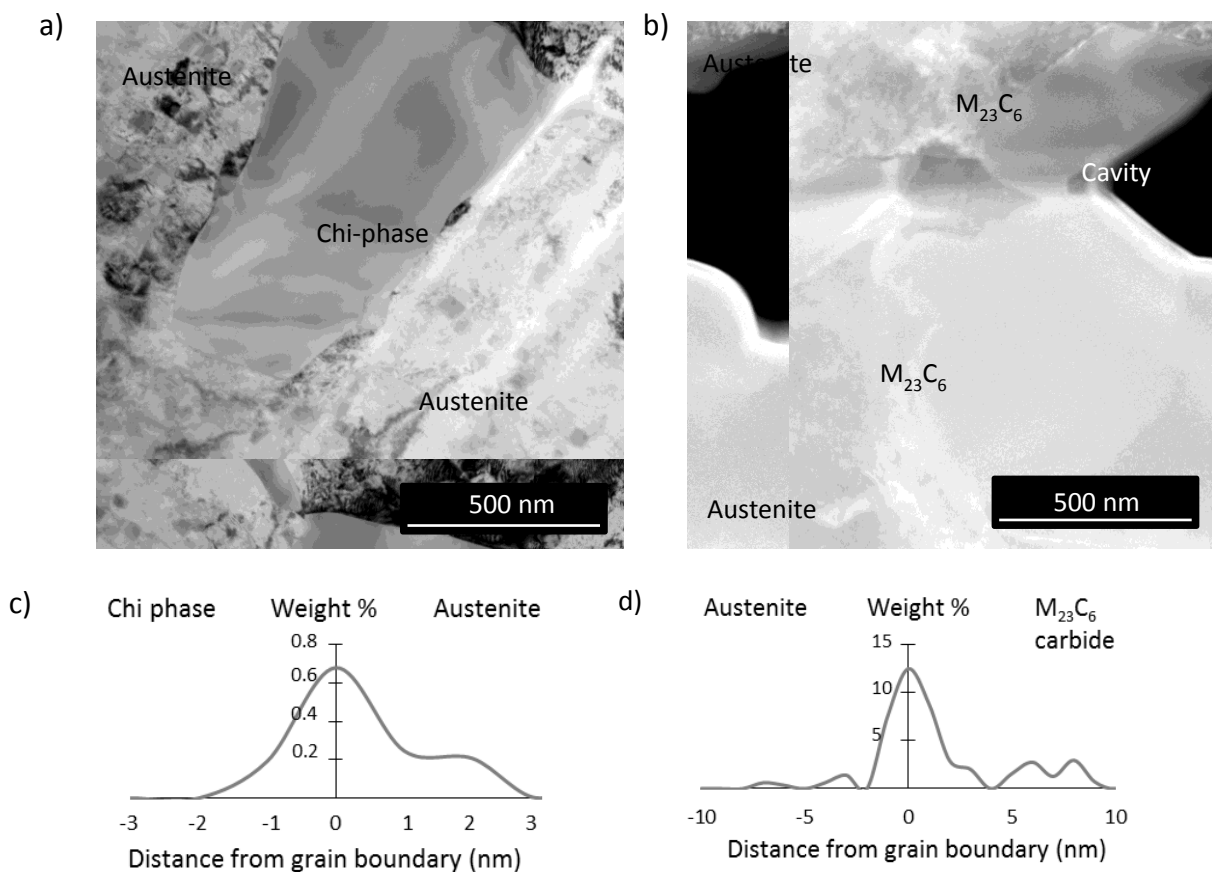


Figure 2: a) TEM image showing the distribution of precipitates in Cr rich region of the ex-service +22000 hour aged Type 316H steel boiler header, b) TEM image showing the distribution of precipitates associated to creep cavity of the bifurcation, c) EDX line across an austenite-chi phase measured phosphorus segregation across the interphase boundary in the Type 316H steel boiler header, d) EDX line across an austenite- $M_{23}C_6$ measured phosphorus segregation interphase boundary in the Type 316H steel tubing.

Concluding comments

The different prior thermo mechanical fabrication work for the header and the tube result in different sources of aged microstructures in the components. This is reflected not just in the secondary phases present in the materials, but in the phases associated with phosphorus segregation. Phosphorus has been observed to modify the precipitation behaviour of $M_{23}C_6$ carbides [8,9] and significantly influences creep fracture, although the exact nature of these effects is contested. Studies have suggested that creep cavity nucleation is enhanced, whilst the growth is potentially retarded [8,9]. Chen observed phosphorus at fracture surfaces of the ex-service boiler header [8], indicating that the chi phase precipitates played a role in fracture, whilst previous workers have suggested that the modifications of creep behaviour are related to changes in the precipitation of $M_{23}C_6$ carbide [10]. In both, the header and the bifurcation tube the creep cavitation is promoted by the phosphorus segregation at the chi-austenite interphase boundary and for the bifurcation tube the carbide-austenite boundary. In both cases the nucleation of creep cavitation will be promoted, controlled by the relationship $r = 2\gamma/\sigma$ where r is cavity radius, γ is interfacial energy and σ is local tensile stress normal to the grain boundary. Therefore on each case, the segregation of the P will further decrease the interfacial energy and make the critical size for cavity nucleation smaller.

Fabrication processes can significantly modify creep behaviour in components, through controlling the initial microstructure and subsequent phase evolution.

Acknowledgements

The authors want to thank EDF Energy for funding this research and for provision the sample materials. We also would like to thank EPSRC for founding South of England Analytical Electron Microscope under grant no. EP/K040375/1.

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