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Ductile Fracture Assessment of 304L Stainless Steel Using 3D X-ray Computed Tomography

A. J. Cooper, O. G. C. Tuck, T. L. Burnett and A. H. Sherry

Abstract Stainless steel manufactured by hot isostatic pressing (HIP) has been shown to exhibit significant differences in ductile fracture behavior when compared to equivalently graded forged steel, due to differences in oxide particle concentration between the two manufacture routes. Herein we analyse and quantify the ductile damage characteristics in the fracture process zone of equivalently graded forged and HIP 304L steel using 3D X-ray computed tomography (CT). Ductile void characteristics have been found to vary in size, shape, and spatial distribution; data which are in agreement with the differences in distribution of initiation particles in HIP and forged steel. Using advanced X-ray CT to characterize ductile damage, experimentally determined data can be employed to calibrate existing well-known ductile failure models, developing both our current understanding of ductile failure as well as a predictive tool to simulate fracture in novel HIP components.

Keywords Hot isostatic pressing • Ductile fracture • X-ray computed tomography • Stainless steel • Finite element modelling

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Introduction

Near net shape (NNS) manufacture is an emerging route for producing structural components used in the nuclear industry. Of these, hot isostatic pressing (HIP), the consolidation of an alloy powder in a mould under high temperature and pressure, has been shown to be effective for producing complex geometry components without the need for welds; thus reducing costs, accelerating manufacturing time, and improving performance [1–4].

One of the fundamental differences between equivalently graded forged and HIP steels lies in the measured oxygen concentrations in the bulk material, with HIP steels typically exhibiting oxygen concentrations over an order of magnitude greater than those found in equivalently graded forged steels. This increase in oxygen content is thought to arise as a result of metal powder surface oxidation [5] during powder storage and handling, and the oxygen manifests itself in the form of non-metallic oxide inclusions, evenly distributed throughout the HIP'd microstructure. It has been proposed that the finer distribution of oxide particles facilitates the ductile fracture mechanism by acting as additional initiating sites for void nucleation and growth [6–8]; the initial stages of the classical ductile fracture mechanism. This results in reducing the distance over which ductile damage (in the form of voids) is required to grow before resulting in void coalescence with neighbouring voids, thus permitting fracture at smaller levels of plastic strain, and causing an unzipping effect throughout the material microstructure. This difference in ductile failure mechanism is reflected in the measure fracture toughness values ($J_{0.2BL}$) for these the two materials; 1000 kJ m^{-2} and 1750 kJ m^{-2} (reported previously [9]) for HIP304L and F304L, respectively.

Although correlations have been established linking oxygen concentration, oxide particle area fraction on metallographic sections, Charpy impact toughness, and differences in ductile dimple characteristics, thus allowing for the development of this working hypothesis [9–12], it has so far not been possible to obtain 'statistically significant' information regarding the ductile void growth characteristics during failure, on which the developed understanding hinges. SEM offers high resolution to probe ductile void characteristics below the fracture surface of sectioned test specimens, however it can be challenging and inefficient since fractured test specimens often exhibit very few voids on each metallographic plane (shown later); analysis therefore requires numerous sections to be prepared and analysed in order to build up an accurate representation of global damage.

In contrast, three-dimensional X-ray computed tomography (CT) offers the ability to study ductile damage in significantly greater detail than has hitherto been observed experimentally, probing the internal void structure of materials and providing quantitative data on the shape, size, spatial distribution, and volume fraction of voids formed below the fracture surface during ductile failure. Importantly, X-ray CT allows for the collection of data with 'statistical significance', whereby imaged specimens may contain thousands of measureable voids that have been subjected to a known stress state and/or loading conditions. In recent years,

X-ray CT has been employed as a tool to study microvoid coalescence in a variety of alloy systems [13, 14], and quantitative X-ray CT has been shown [15–17] to be a promising tool for the determination of ductile damage parameters to inform the predictive model for ductile failure developed by Gurson, Tvergaard and Needleman [18–20].

The Gurson-Tvergaard-Needleman (GTN) model is one of the more popular classical approaches used to predict ductile failure in materials. It is based on a yield criterion for a porous medium, which assumes the influence of damage (in the form of voids) on plastic flow can be accounted for by averaging the behavior within a discrete material volume element using a homogeneous continuum mechanics approach. The model accounts not only for material strain hardening but also the softening that occurs as a result of the nucleation and growth of voids during plastic flow. The model is able to predict material failure, which is assumed to occur when a critical volume fraction of voids is attained within the volume element. However, the implementation of the GTN model is not straightforward; it contains a number of parameters, some of which are interdependent, that must be appropriately calibrated for the material of interest. These parameters are assumed to be material constants, relating to the microstructural characteristics and nature of damage accumulation; once calibrated against a particular dataset can be used to predict ductile failure in other geometries under different loading conditions. Typically, calibration of the GTN model is achieved iteratively, by estimating parameter values using the results of previous studies, performing a Finite Element Analysis (FEA) of laboratory fracture mechanics tests using these parameters, and comparing the simulation results to the experiment, and adjusting the parameters accordingly. This can be very time consuming, and the varying methods used to adjust the parameters by different researchers can lead to very different parameters for the same material and test conditions [21, 22]. Furthermore, the underlying physical processes involved do not guide the adjustment of parameters; there is a danger that values of the calibrated parameters bare little correspondence to the physical quantities they are meant to represent. Quantitative microstructural analyses are therefore needed to better inform the setting of parameters. Hence, not only can XCT studies improve our comprehension of ductile fracture by providing qualitative information regarding failure mechanisms, but are also able to deliver quantitative data to support the calibration of predictive models.

In this paper, the ductile fracture behaviour of 304L stainless steel is examined using X-ray CT, and we explore various avenues through which the GTN model could be calibrated using the experimental data.

The three-dimensional tomography scans were performed on an instrument capable of resolving voids of only a few microns in size, allowing for an accurate mapping of void populations without the need for laborious metallographic analyses as well as allowing for the analysis of a much larger sample size. Furthermore, the results of the X-ray CT were processed to generate quantitative data for distributions in void size and spacing.

This work expands on previous work, providing additional experimental evidence to highlight the differences in mechanistic ductile fracture between HIP and

forged variants of 304L stainless steel, and employs the use of SEM to study the differences in microstructural damage ahead of and surrounding the crack tip of sectioned failed test specimens. This work also expands on the work by Daly et al. [15, 16], who employed the use of 3D X-ray CT to probe ductile damage in SA508 weld metal in order to improve the calibration of the Gurson, Tvergaard and Needleman [18–20] ductile fracture model, by investigating approaches to approximating the fracture surface of cores which exhibit a large depth of fracture process zone.

Background: Continuum Damage Mechanics Models

The GTN model is an example of a continuum damage mechanics model in which the influence of an initial volume fraction of porosity (f_0), the nucleation and growth of said porosity as well as newly initiated voids, and subsequent coalescence of voids on the load-bearing properties of material elements is simulated. The model is integrated into the constitutive equation of material behaviour within FEA such that the plastic flow potential (Φ) is given by:

$$\Phi = \left(\frac{\sigma_{eq}}{\sigma_f} \right)^2 + 2q_1 f \cosh \left(\frac{3q_2 \sigma_h}{2\sigma_f} \right) - \left(1 + (q_1 f)^2 \right) \quad (1)$$

where σ_{eq} denotes the von Mises equivalent stress, σ_f denotes the matrix flow stress, σ_h denotes the hydrostatic stress and q_1 and q_2 denote material parameters first introduced by Tvergaard who defined $q_1 = 1.5$ and $q_2 = 1.0$ [18, 19].

The void volume fraction f is calculated from the growth rate of initial porosity, $\dot{f}_{gr}$, and the nucleation rate of newly initiated voids, $\dot{f}_{nu}$. The growth rate of the total void volume fraction is derived from the plastic part of the strain rate tensor. The nucleation rate of new voids is related to the equivalent plastic strain rate

$$\dot{f} = \dot{f}_{gr} + \dot{f}_{nu} = (1 - f) \dot{\epsilon}_{ii}^p + A \dot{\epsilon}_{eq}^p \quad (2)$$

where

$$A = \frac{f_N}{S_N \sqrt{2\pi}} \exp \left[-\frac{1}{2} \left(\frac{\epsilon_{eq}^p - \epsilon_N}{S_N} \right)^2 \right] \quad (3)$$

where the parameter f_N denotes the volume fraction of particles that initiate new voids and ϵ_{eq}^p is the equivalent plastic strain. The parameters ϵ_N and S_N define a normal distribution of nucleation strain with ϵ_N being the mean strain and S_N the standard deviation.

Tvergaard and Needleman [20] included a rapid increase in void volume fraction at high plastic strains to simulate the coalescence phase in the ductile fracture

mechanism. They added a parameter f^* to replace f so that above a critical value, denoted f_c , there is a rapid increase in f^* to account for coalescence and the rapid softening of the material just prior to failure. Below this critical value, f^* is equal to f defined by Eq. (2), i.e.

$$f^*(f) = \begin{cases} f & \text{for } f \leq f_c \\ f_c + \frac{f_u^* - f_c}{f_F - f_c} (f - f_c) & \text{for } f > f_c \end{cases} \quad (4)$$

The void volume fraction f_F is the value of f^* at which final fracture is deemed to have occurred. The parameter $f_u^* = 1/q_1$.

In summary, the primary parameters that require calibration are the following: f_0 , f_c , f_F (that describe void growth and coalescence to failure), ϵ_N , s_N , f_N (that describe void nucleation), q_1 and q_2 . Since GTN models are solved using FEA, another parameter, associated with the crack-tip mesh size within the model must also be determined since results are dependent on the resolution of stress and strain fields at the crack-tip. This parameter, the crack-tip mesh size, denoted L_c , is related to the spacing of the large oxide inclusions, since the most prominent voids observed using tomography are likely to initiate at these large oxide particles.

Experimental

Material

The material examined was hot isostatically pressed 304L material, supplied by Areva, and forged 304L stainless steel by Creusot Forge et Creusot Mécanique, Areva (Le Creusot, France). For HIP304L, stainless steel grade 304L powder was heated from ambient temperature to 1123 K (1150 °C) at a rate of 633 K (360 °C) h^{-1} and held at 1123 K (1150 °C) and 104 MPa for a period of 180 min. Post-HIP heat treatment of HIP304L was performed by heating from room temperature to 1343 K (1070 °C) at 633 K (360 °C) h^{-1} , held for 280 min, and water quenched. Forged 304L pipe was subjected to similar heat treatment as the HIP materials (1343 K (1070 °C), for *ca.* 250 min) and water quenched.

The materials' elemental compositions (weight %) and grain sizes are tabulated in Table 1. Grain size measurements were conducted in accordance with ASTM E112-96. [23]

J-R fracture toughness testing was performed on F304L and HIP304L, the experimental details and results of which are presented in a previous study [11]. Specimens selected from the fracture toughness testing for analysis herein were loaded to an applied J and crack growth of 742 kJ m^{-2} /0.454 mm and 672 kJ m^{-2} /0.367 mm, for HIP304L and F304L, respectively. These specimens were selected as they exhibited the smallest degree of crack growth and loaded to the smallest

Table 1 Elemental composition

		Grain size	Cr	Ni	Mo	Mn	Si	C	O/ ppm	N/ ppm
304L	Spec. (%wt.)	–	18.5–20.00	9.00–10.00	–	<2.00	<1.00	<0.035	200	–
	Forged	94 μm	19.40	9.648	0.345	1.654	0.573	0.027	15	817
	Powder (%wt.)	–	19.2	9.44	–	1.37	0.74	0.022	110	–
	HIP (%wt.)	27 μm	19.5	9.45	0.01	1.33	0.72	0.022	120	840

applied J , in an attempt to assess specimens which were loaded to within the J_{max} validity limit of 738 kJ m^{-2} , as determined using ASTM E1820 [24].

Specimens prepared for metallurgical analysis were sectioned, mounted, ground, and polished in accordance with the recommended procedures in ASTM practice E3-01 [12].

Electron microscopy was performed using an FEI Quanta 650 ESEM and an FEI Sirion SEM, both equipped with field emission guns and Electron Back-Scattered Diffraction (EBSD) detectors, and a Hitachi S-3700 scanning electron microscope equipped with Oxford Instruments INCA X-ACT energy dispersive spectroscopy for semi-quantitative chemical analysis. The SEM was performed under vacuum using a 20 kV accelerating voltage and a spot size of 4.0 nm, at a working distance (WD) of approximately 10 mm.

Tomography Specimen Preparation

Cylindrical core specimens were extracted from failed C(T) specimens (HIP304L and F304L) using electro-discharge machining (EDM). The samples measured approximately 0.5 mm in diameter and 20 mm in length. Three cores were extracted along the ductile tearing zone of each C(T) specimen, and a single core was extracted in a region of the machined notch far from any crack propagation; this was taken in order to detect the presence of any initial voids within the material in order to distinguish between initial porosity and ductile damage. Extraction of the cores from the ductile tearing region was located as central to the specimen as reasonable possible, such that the material under investigation had been subjected to plane strain conditions.

The extraction plan for the cores is shown in Fig. 1.

Fig. 1 EDM sample extraction plan. Three samples were extracted for each specimen within the ductile tearing zone (Left (L), Centre (C) and Right (R)) of a failed CT half specimen. A fourth sample was extracted behind the fatigue pre-crack away from any ductile damage

3D X-Ray Tomography

3D rendered X-ray CT was performed on each core using a Zeiss Versa 500 machine with a 4X objective, using 100 kV X-rays and 9 W power. An exposure time of 8 s was used, and an LE3 filter was employed. A final voxel size of $0.75 \mu\text{m}^3$ was obtained, for the 4X scans.

Data processing was performed using Avizo[®] 9.0.0 software. In order to reduce image noise and improve contrast between voids and metal, an edge-preserving smoothing filter was applied to the data. The immediate surface perimeter of the core was cropped by five voxels to remove the very porous surface as a result of EDM machining of the cores, which may contribute to the inaccurate rendering of voids. The procedure used to segment voids and metal in the X-ray CT reconstructed images is detailed by Daly et al. [16]; to which the reader is directed for details. Void segmentation was performed such that any voids in any contact with the fracture surface were assigned to the air surrounding the core. This was to ensure consistency when segmenting voids close to/in contact with the fracture surface. In order to quantify size, shape, and nearest neighbour distances of voids, the spatial coordinates and volumes of each void were extracted. In order to manipulate the data usefully, it is useful to know the spatial location of damage relative to the fracture surface; various methods were explored for this and are detailed in the results section.

Void volume fractions were calculated as a function of distance from the fracture surface, and was determined by calculating the volume fraction of each reconstructed 'slice', where a slice had a thickness equal to that of the voxel resolution, e.g. 0.75 μm .

Results

Table 1 shows the elemental composition of the two materials studied, the specification for 304L, as stated on the material certificate for 304L, is also tabulated. Both of the materials are within specification. The oxygen concentration is not usually specified on materials certificates; however it is clear that the oxygen content in the HIP304L is over an order of magnitude greater than that of F304L. This observation was also made in previous stainless steel studies [9–11].

Figure 2 shows reconstructed CT scans of cores extracted from regions far away from the ductile tearing zone, for (a) F304L and (b) HIP 304L. In both specimens, there appears to be a small volume fraction of particles that are above the 0.75 μm resolution of the CT scans. Because no porosity has previously been observed in these materials [10, 11], it is believed that the particles are non-metallic oxide inclusions. 57 and 74 inclusions were resolved in the HIP304L core and F304L core, respectively. The average radii of the oxide inclusions in the cores were measured to be 2.86 μm and 2.67 μm , for HIP304L and F304L, respectively. It is important to note that there will be a appreciable number of inclusions in the microstructure that are smaller than the resolution of the CT scanner and objective for this particular experimental setup; indeed previous microstructural analyses of similar HIP 316L materials revealed oxide inclusions with radii on the order of 0.3 μm [11], which would be below the resolution of these CT scans.

Figures 3 and 4 show reconstructed CT scans of F304L and HIP304L, respectively, and ductile damage is observed below the fracture surface in each core cores (highlighted blue). It can be seen in Fig. 2 that the voids in the F304L cores display slight elongation in the direction of maximum principal plastic strain, which is not as obvious in the HIP304L specimens. It is thought that this is related to the degree of large scale plasticity surrounding the crack tip, which is significantly more prominent in the F304L cores than HIP304L, and which results in a more heavily deformed crack tip region due to excessive levels of plastic strain. Evidence of much more prominent plasticity in F304L is also shown in the topography of the cores, whereby F304L cores exhibit a fracture surface on a sharp gradient, in contrast to HIP304L which exhibit core topography with comparatively flatter fracture surfaces.

In contrast, the damage below the fracture surfaces of the HIP cores appears largely spherical, with minimal evidence of elongation. Even by eye, there appears to be a greater degree of ductile damage in the H304L cores than in the F304L cores, and damage is localised to a region of approximately 250 μm below the fracture surface.

Fig. 2 3D rendered EDM cores extracted from the loading pin region of (a) F304L, and (b) HIP304L, showing a small volume fraction of initial particles in the microstructure

Further evidence of F304L's higher crack tip deformation is shown in Figs. 5 and 6, which show backscatter electron (BSE) images of sectioned compact tension test specimens, and highlight the various stages of crack growth; moving from right to left, fatigue pre-crack, controlled ductile tearing zone (from which the X-ray CT cores were extracted), followed by the end of ductile tearing and beginning of fatigue post-crack. The BSE images highlight the degree of microstructural damage ahead of and surrounding the crack tip, and a clear definable region of large plastic deformation (plastic zone) can be seen at the crack tip of F304L (Fig. 5) with an approximate radius of *ca.* 500 μm . In contrast, the plastic zone is much more difficult to distinguish in HIP 304L (Fig. 6), partially due to the finer grain size, but also due to significant less crack tip blunting. In both sectioned specimens (Figs. 5 and 6),

Fig. 3 3D rendered EDM cores extracted from specimen F304L showing a side view (fracture surface at the top) of the reconstructed 4X X-ray CT data

there are clear and significant levels of plastic slip, identifiable by the slip lines orientated within grains.

It can be seen at the crack tip of F304L that there is a region where the material resembles complete plastic flow, whereby individual grains are no longer identifiable, thus indicating the high levels of crack tip blunting which has occurred during loading. This is not as identifiable in HIP304L.

It is interesting to note how few voids are visible in the sectioned BSE images in comparison to the 3D rendered X-ray CT images, which emphasises the 3D nature of the ductile void damage and highlights how sectioning specimens to assess the extent of ductile void damage can be both challenging and misleading. This is not surprising since metallographic sectioning of specimens only reveals ductile damage that is located on the sectioned plane, and it is not possible to determine void sizes accurately since sectioning does not necessarily, and is very unlikely to, occur through the centre of the void. This results in voids appearing smaller than the actual radii they possess.

Fig. 4 3D rendered EDM cores extracted from specimen HIP304L showing a side view (fracture surface at the top) of the reconstructed X-ray CT data

Characterising Ductile Damage

The average VVF data for HIP304L and F304L, as presented in Fig. 7, indicate very little difference between the two materials when considering damage represented as VVF; interesting given the materials' appreciable difference in fracture toughness.

It is challenging to quantify any differences in ductile damage of HIP and Forged materials, when representing damage in the form of void volume fractions as a distance below the fracture surface. This is perhaps a reasonable result, since it could be argued that in theory, the two materials should exhibit the same *volume fraction* of damage at failure, since the HIP material exhibits more, finer voids, whereas the Forged material exhibits fewer, larger voids.

This argument is further strengthened when characterizing the average void sizes and the extent of ductile damage in terms of total 'void count', which represents the total number of resolved voids in the material. Total void count has been measured and plotted as a function of distance below the fracture surface, in bins of 50 μm thickness, for HIP304L (Fig. 8) and F304L (Fig. 9). The most notable observation is the much larger total void count for the HIP cores when compared to the Forged cores; the total number of voids observed in the three F304L cores was 1232 whereas the total number of voids resolved in the HIP 304L cores was 5042, an increase in over 300%.

Fig. 6 BSE images of the crack tip of HIP304L showing the various stages of the fracture process, and the ductile tearing region from which cores were extracted, as well as evidencing various damage features; slip bands, material plastic flow, and ductile voids

Fig. 7 Average VVF fraction data for HIP304L (blue) and F304L (black) as a function of distance below the core tip

Fig. 8 Total void count of the F304L cores presented as a histogram with bin sizes of 50 μm as a function of distance below the core tip

Fig. 9 Total void count of the F304L cores presented as a histogram with bin sizes of 50 μm as a function of distance below the core tip

Discussion

Through using X-ray CT to visualise the extent of ductile damage in fractured compact tension test specimens, various damage characteristics can be extracted and analysed. It has been shown previously that equivalently graded HIP and Forged stainless steel exhibits clear differences in fracture and impact toughness, as well as the number of oxide particles in the microstructure, which act as initiation sites for the nucleation and growth of voids during ductile failure. Whilst the spatial distribution of these initiating particles varies significantly between HIP and forged 304L, the stress strain behaviour of the materials has been previously reported as being comparable exhibiting similar strain hardening behaviour, which indicates that the rate of void growth should be similar for specimens tested under equivalent loading conditions. The rate of void growth can be estimated using various void growth models, such as those proposed by Rice and Tracey [25] and McClintock [17], and indeed Rice and Tracey calculations have been shown to glean comparable results for both HIP and forged 304L [9]. This is significant since it suggests that the ductile fracture mechanism of these two different steels is not governed by the void volume fraction of voids below the fracture surface, and instead is dominated by the inter-particle distance of nucleating voids. This is perhaps intuitive, since the total volume fraction of fewer, larger voids (as seen in the forged specimens) would be expected to equate to the total volume fraction of more, finer voids, at void coalescence and fracture initiation, if said voids are assumed to grow with comparable growth rates.

Indeed, the coalescence stage of failure will occur at different levels of plastic strain for two materials exhibiting different average inter-particle spacing, but the VVF of ductile damage below the fracture surface at each material's respective fracture initiation, providing the materials exhibit similar strain hardening behaviour, should be comparable. Though this has formed the basis of a working hypothesis for recent works, this is the first set of experimental data evidencing how the ductile damage evolves between the two materials during failure. Significantly, if little differences between VVF data are discernible, then significant differences between the void size, distribution, and frequency should be observable in order to explain the differences in fracture and impact toughness previously reported. The total number of resolvable voids was found to be 5042 for HIP and 1232 for F304L, an increase in over 300%. The average void radii were found to be 3.58 μm for HIP304L and 5.46 μm for F304L; a 53% increase in average void size for F304L. Voids in HIP304L were found to be more finely spaced than in F304L, with a reduction in distance on average of 16.34 μm . These significant differences in ductile damage characteristics are representative of the difference in fracture toughness between the two materials.

In principle, it is well within the capability of X-ray CT to accurately characterise the damage that accumulates during ductile failure. It has been shown here that void volume fractions, nearest neighbour distances, and void shapes and sizes, which are all characteristic of ductile failure, can be measured in 3D using X-ray

imaging, and these data can be employed in continuum damage mechanics models such as the GTN model. The GTN model in particular employs the use of void volume fractions to quantify damage at various stages of failure (f_n , f_c , and f_F) as well as an additional void volume fraction parameter associated with the initial porosity in the material (f_0). The challenge here is to assign particular void volume fractions to the various stages of ductile failure (void nucleation, growth, and coalescence) which might be possible through the use of nearest neighbour distance and void shape characteristics. The specimen cores analysed in this work can be thought of as capturing the failure history of the test specimen. Voids that are located remote from the fracture surface (and crack tip), and which reside towards the limit of the plastic zone can be associated with void nucleation, rather than growth and coalescence. These voids are not subjected to sufficiently high plastic strains such as to cause necking and coalescence with neighbouring voids, because they do not have the opportunity due to the failure process at the crack tip. Therefore it might be plausible to assume a region remote from the fracture surface, within which there average void volume fraction can be associated with f_n . Similarly, in regions close to the fracture surface, voids will experience levels of plastic strain sufficiently large enough to result in coalescence with neighbouring voids. These voids, as a result of the necking process, are expected to exhibit non-sphericalness, and it might be possible to assign a probability of coalescence based on void geometry. Finally the void volume fraction of damage associated with failure (f_F) may be characterised using the measured void volume in direct contact with the fracture surface of the cores. The void volume fraction associated with initial porosity (f_0) is associated with porosity (or voids) in the initial unloaded microstructure (i.e. inclusions at which voids form). As shown previously, X-ray CT might not be the most appropriate method to determine this parameter, since the vast majority of non-metallic inclusions in the microstructure will have radii, which are below the resolution of even the most advanced high-resolution CT scanners. Scanning electron microscopy on several sections of material, or indeed through the use of Plasma focused ion beam (FIB), would be the most appropriate methodology to obtain accurate void volume fraction data associated with f_0 .

Clearly there are several approaches that could be employed here to separate and assign damage to the various stages of ductile failure, which merits its own separate investigation and sensitivity study. As well as the various stages of failure, the GTN model assumes an inter-particle spacing (implemented into an FE model via the mesh size). This inter-particle spacing can be assumed to be the nearest neighbour distance of voids observed using X-ray CT. It is important to realise here that voids close the fracture surface (and crack tip) will experience sufficiently large plastic strains to grow to sizes detectable by the CT scanner, and thus, decay in number of voids (and increase in nearest neighbour distance) observed as a function of distance below the fracture surface is a result of the stress field (and plastic zone). Because of this, it can be assumed that the small nearest neighbour distance values in close proximity to the fracture surface is actually constant and extends throughout the core. This nearest neighbour distance of voids can be usefully employed as the inter-particle spacing (and mesh size) and would play an important

role in governing the fracture toughness of ductile materials. Significantly, it has been shown here that there are little discernible differences between the damage characteristics when considering void volume fractions as a function of distance from the fracture surface (Fig. 7). By extension of this, it might be expected that the two materials would also exhibit comparable void volume fractions for the various stages of failure, and instead, the fracture mechanism is governed by the inter-particle spacing (nearest neighbour distance), which differs on average by *ca.* 16 μm .

Work remains to accurately assign observed damage to the various stages of failure and develop this approach to calibrate such ductile failure models, but the authors hope to highlight the capabilities of this form of experimental analysis to better understand the mechanisms of ductile failure and by extension, to accurately calibrate existing ductile failure models.

Conclusions

This work has used 3D X-ray computed tomography to probe the fracture behaviour of equivalently graded HIP and forged 304L stainless steel, and for the first time, has revealed the differences in ductile void characteristics in these materials when subjected to severe loading conditions. The work concludes recent investigations into the mechanistic failure of HIP steels when compared to forged steels, and confirms a working hypothesis regarding how fracture toughness is governed by the spatial distribution of non-metallic oxide inclusions, which act as initiation sites for the nucleation, growth, and, crucially, the coalescence of neighbouring voids during ductile failure. The authors also propose the appropriateness of X-ray CT for determination of relevant parameters required to calibrate the GTN continuum damage mechanics model.

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