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FRACTURE SURFACE MORPHOLOGY UNDER DUCTILE TEARING OF METAL PLATES

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Key words: Ductile plate tearing, Microstructure randomness, Void-by-Void, Multiple void interaction, Localization.

Summary. The present work takes as offset the hypothesis that microstructural parameters, related to particle size and distribution, govern the transition between crack surface morphologies observed in experiments. The key question is; why does tearing of a given metal plate leave a specific morphology on the fracture surface?

1 INTRODUCTION

The micro-mechanisms governing ductile plate tearing have been established through decades of research. Fracture of ductile metal plates, subject to mode I tearing, is known to depend heavily on the metal properties, microstructure and plate dimensions. Intermetallic inclusions are often responsible for the nucleation of the micro-voids that, upon loading, grow to coalesce and eventually form micro-cracks. However, distinct crack surface morphologies, such as cup-cup, cup-cone, and slanting are observed for extensive crack growth in metal plates and several investigators report that plates made of high strength age-hardened aluminum alloys, or high strength steels, exhibit slanted crack growth, whereas a cup-cup type fracture is observed for low strength alloys. Despite such insight, the interplay between tearing modes remains to be fully understood, e.g. nowhere is a conclusive answer found to; *why does a propagating crack choose one tearing mode over others?* In the present, work focus is turned to the effect of microstructure - in terms of void nucleation sites, number and distributions. The hypothesis is that; a population of small particle/void nucleation sites, distributed randomly in the material, gives rise to a very different fracture behavior in comparison to a population of randomly distributed large particle/void nucleation sites with the same overall volume fraction. Essentially, a shift in the governing mechanism from multiple void interactions, in a macroscopic deformation band, to a microscopic localization process that occurs as void-by-void interaction, is expected.

2 MODELLING FRAMEWORK

The present work adopts, and modifies, the simplified 2D plane strain model setup considered by Nielsen and Hutchinson (2012). Rather than restricting the analysis to a homogeneous Gurson-Tvergaard-Needleman (GTN) material (a continuum description), the present work focuses on the effect of introducing a population of “discrete” particles that act as discrete void nucleation sites. Thus, the model setup allows microstructural parameters such as the average initial porosity, the particle size (relative to plate thickness), number and distribution characteristics to enter the analysis. A number of void nucleation sites, N_p , are considered to be distributed randomly and modeled by locally introducing a high level of damage that locally follows a Gaussian bell (see Eq. (1)). Here, with the constraint that no overlapping between neighboring nucleation sites are allowed.

$$f_0(x_1, x_2) = f_0^M \exp\left(-\frac{(x_1 - x_1^M)^2 - (x_2 - x_2^M)^2}{2(R_p/4)^2}\right), \quad (1)$$

with f_0^M being the porosity at center of the nucleation site, x_i^M are the coordinates of the nucleation site center, and R_p is the radius of the nucleation site (with a minimum of 6 elements across each nucleation site). A background porosity is assumed to exist between nucleation sites, f_0^b , which allows neighboring sites to eventually coalesce. The GTN material model is employed to predict the porosity evolving from the nucleation sites as-well as the macroscopic material separation that develops across the plate thickness as voids link up. The employed version of the GTN model reads;

$$\Phi = \left(\frac{\sigma_e}{\sigma_y}\right)^2 + 2q_1 f^* \cosh\left(\frac{3}{2} \frac{q_2 \sigma_{kk}}{\sigma_e}\right) - (1 + (q_1 f^*)^2), \quad (2)$$

and takes into account void coalescence by the phenomenological f^* -criterion introduced by Tvergaard and Needleman (1984). Thus, the analysis relies on two key void volume fractions; one to define the onset of void coalescence, f_c , and one to define the complete loss of load carrying capacity at a material point, $f_f (= f_0^M)$. The Tvergaard constants are; $q_1 = 1.5$ and $q_2 = 1.0$.

As plastic flow localizes ahead of the crack tip, the domain above and below the fracture process zone elastically unload. This enforces a constraint on the deformation along the x_3 -direction that leads to a near tip plane strain condition (see Fig. 1). That is, a cross-sectional plane, assumed constraint to plane strain in the crack growth direction, is modeled by use of ABAQUS/Explicit and the section considered maintains an initial aspect ratio of $H/W = 3$. Figure 1 outlines the boundary value problem considered. The element size in the finely discretized section is; $L_e/W = 0.001$, with W being the plate thickness. Elements in that region are 4-node and square in the un-deformed configuration (employing reduced Gauss integration, CPE4R). ABAQUS/Explicit is employed to facilitate element erosion when a sufficient damage level is reached, but it is not the intention to enter a regime where inertia plays a role.

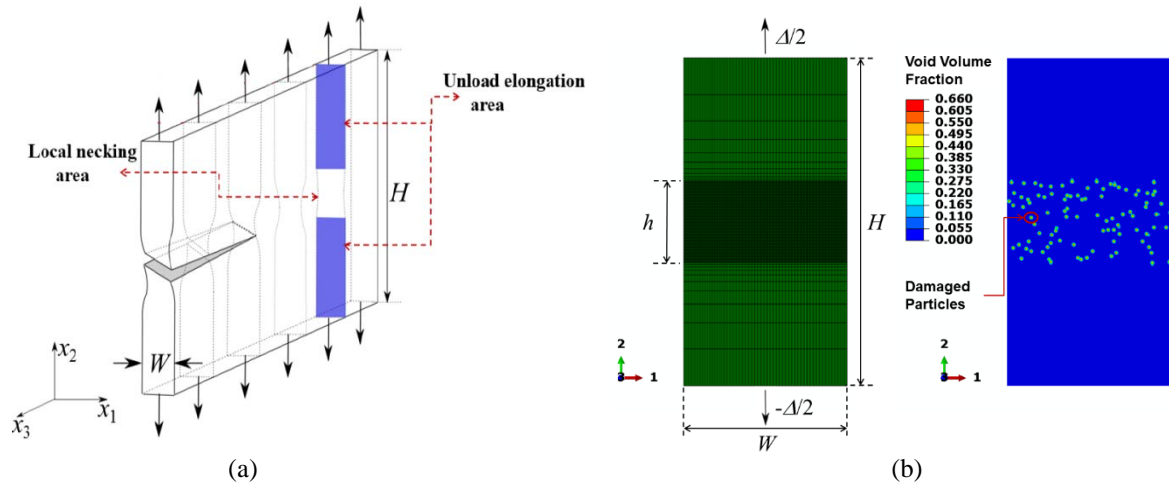


Figure 1: (a) Schematic of the sequential fracture process governing ductile plate tearing under mode I loading, here identifying the domain height, H , and the plate thickness, W . (b) The FE mesh for the 2D plane strain plate section with imposed boundary conditions and illustrated damaged particle distribution.

3 RESULTS

The plate material is assumed to be a typical aluminum alloy in the calculations presented below (with a Young's modulus; $E = 70$ GPa, Poisson ratio; $\nu = 0.3$, density; $\rho = 2700$ kg/m³, initial yield stress; $\sigma_0 = 300$ MPa, and a power-law hardening exponent; $n \in [0.05, 0.15]$). An extensive convergence study has been performed to adjust the model parameters; f_c , f_f , f_0^b , and the imposed macroscopic strain rate, \dot{E}_{22} , such that dynamic effects are eliminated from the results. Here, by ensuring a low kinetic energy and a rate-independent crack surface morphology (the latter proving to be a strict condition). The convergence study showed that with increasing initial porosity in the plate (which depends on f_f , f_0^b , and N_p), the \dot{E}_{22} value required to provide quasi-static results decreases. This is in accordance with the physics of the problem: a larger initial porosity leads to several energetically (more or less) equivalent crack paths and even a small change in \dot{E}_{22} results an alternative crack path to be activated. Also, for f_0^b values on the same order of magnitude as f_c , the crack path evolves through void-by-void coalescence and is rather insensitive to \dot{E}_{22} . Figure 2 shows preliminary results and, as expected, the crack path (here depicted in the undeformed geometry) is indeed sensitive to the number, distribution and size of the damaged discretely modelled particles as well as to the strain hardening capacity of the plate.

4 CONCLUSIONS

A numerical framework is developed to investigate the effects of material/geometric parameters on the crack path morphology in ductile tearing of metal alloys. Although more calculations need to be performed to reach secure conclusions, the initial results suggest that a small (large) strain hardening exponent favors a slanted (cup-cup) crack, in accordance with the experimental results in the literature.

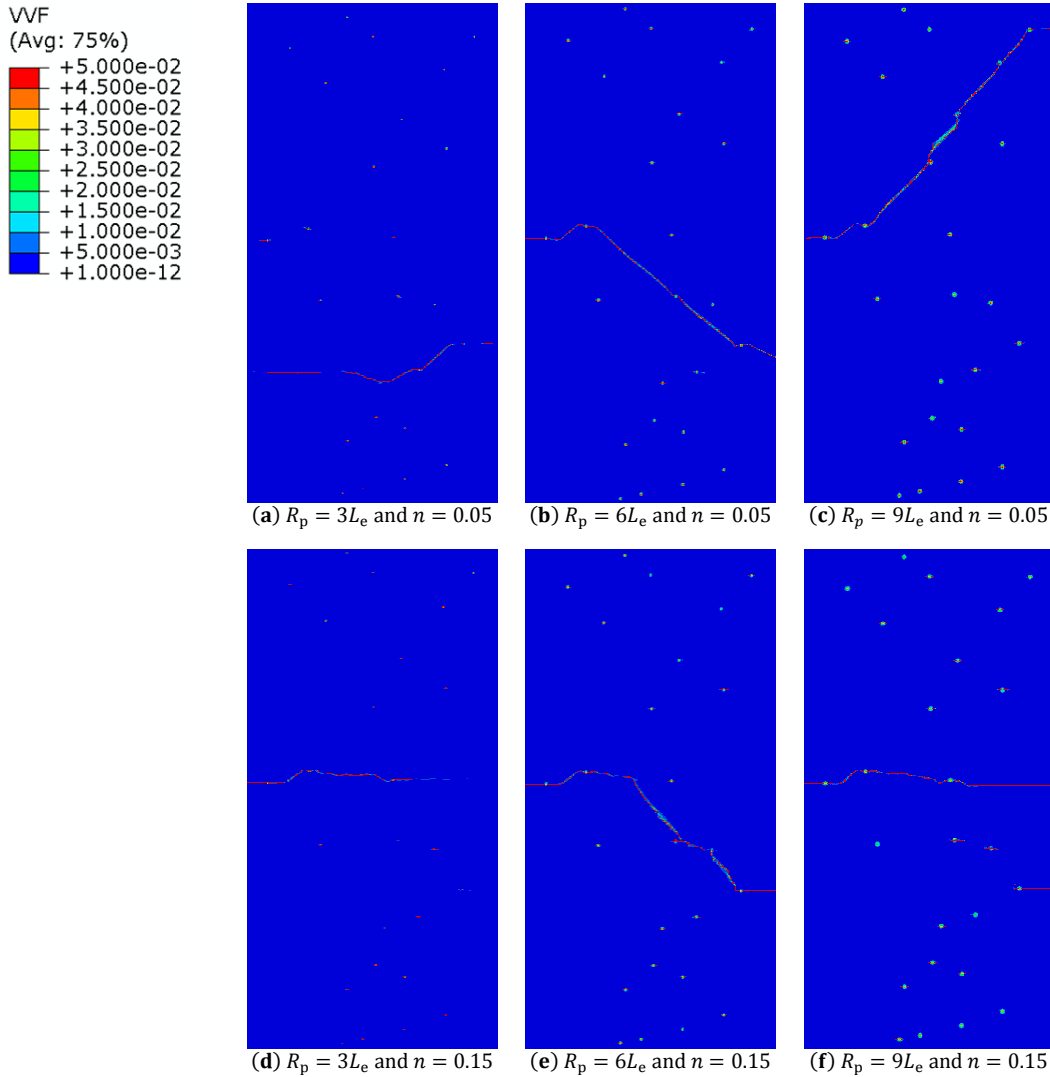


Figure 2: (a-f) Crack path morphologies in the un-deformed geometry for $N_p = 25$, $f_c = 0.01$, $f_f = 0.05$, $f_0^b = 0.0001$, and $\dot{E}_{22} = 0.0008 \text{ s}^{-1}$. The particle radius, R_p , and strain hardening, n , are given below each figure.

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