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# Anisotropic, rate -dependent ductile fracture of Ti-6Al-4V alloy

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## Abstract

An extensive experimental campaign was run to investigate the influence of the loading direction, stress state (triaxiality ratio ranging from -0.5 to 1), and strain rate (from  $10^{-3}$  to  $1.5 \times 10^3 \text{s}^{-1}$ ) on the ductile fracture of Ti-6Al-4V titanium alloy. Microscopic and macroscopic observations provided some insight into the shear-driven or micro-voiding-controlled damage mechanisms prevailing at low and high triaxiality ratios, respectively. Numerical simulations were run to determine the local loading paths to fracture in terms of plastic strain as a function of stress triaxiality ratio and Lode parameter. The ductility was found to be anisotropic, but only weakly dependent on the strain rate in the considered range. The anisotropy in ductility was different in tension (maximum along DD) and in compression (maximum along ND). The fracture strain decreased with the absolute value of the triaxiality, with a maximum close to zero. No clear correlation with the Lode parameter was found.

*Keywords:* Ductile fracture, titanium alloy, triaxiality; Lode parameter, anisotropy; temperature, strain rate

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## 1 Introduction

To be certified some critical parts of aeronautical and space structures need to be designed against accidental overloads to prevent their fracture. This is the case for aircraft engines fan blades that should be able to withstand ballistic events, like bird strike or fan blade loss. During such events the fan blades leading edge – often made of Ti-6Al-4V titanium alloy – undergoes large deformation, high strain rate, non-proportional multiaxial loading and load reversals. The risk of ductile fracture must accordingly be assessed for these complex and widely variable conditions to be further accounted for in the design process. The first stage towards such an assessment was to formulate, calibrate and implement into a commercial finite element computation code (LS-DYNA) a proper constitutive model for Ti-6Al-4V titanium alloy. This stage, detailed in a companion paper, see Ruiz de Sotro et al. (2020), is only briefly recalled here in Appendix since the present paper focuses on damage and fracture.

Experimental works available in the literature, see e.g. Hooputra et al. (2004), Barsoum and Faleskog (2007), Haltom et al. (2013), have evidenced two regimes of ductile fracture of metals and alloys: a high hydrostatic tension regime, in which fracture results from micro-void nucleation, growth and coalescence, and a shear-pressure-dominated regime, in which fracture is controlled by shear strain localization, and decohesion. This loading-dependent competition between void growth-induced fracture and shear decohesion-induced fracture can be modelled

21 by more or less explicitly accounting for the stress triaxiality ratio  $\chi$ , expected to control micro-  
 22 void collapse/growth, and the shear stress ratio  $\eta$ , expected to control micro-void flattening and  
 23 subsequent shear decohesion, or the Lode angle  $\theta$  or parameter  $L = -\cos(3\theta)$ , which  
 24 additionally allows a differentiation of loading types involving similar shear stress states.  
 25 Starting from the definition of the pressure and maximum shear stress  
 26  $p = -\text{Tr} \underline{\underline{\sigma}} / 3 = -(\sigma_1 + \sigma_2 + \sigma_3) / 3$ , and  $\tau_{\max} = (\sigma_1 - \sigma_3) / 2$  with respect to the ranked principal  
 27 stresses  $\sigma_p = \text{eig}_p(\underline{\underline{\sigma}})$  ( $\sigma_1 \geq \sigma_2 \geq \sigma_3$ ), one can express the stress triaxiality and shear stress  
 28 ratios  $\chi = -p / q$  and  $\eta = \tau_{\max} / q$ , where the equivalent stress  $q$  can merely correspond to Von  
 29 Mises stress  $\sigma_M$ , i.e.  $q = \sigma_M$ , or can take a more complex form when e.g. plastic anisotropy  
 30 is involved – as it is the case in the present work. The Lode parameter  $L$  which is defined as  
 31  $L = [2\sigma_2 - (\sigma_1 + \sigma_3)] / (\sigma_1 - \sigma_3)$  can accordingly be rewritten as  
 32  $L = (3/2) \left( [2\sigma_2 / (\sigma_1 - \sigma_3)] - (\chi / \eta) \right)$ .  
 33

34 Ductile fracture criteria may be derived from continuum damage mechanics (CDM)-based  
 35 constitutive models (involving the micro-void volume fraction  $f$ , or more generally a damage  
 36 internal variable  $D$ ), or directly written in terms of critical strain at fracture  $\varepsilon_F^p$  or stress at  
 37 fracture  $\sigma_F$ .

38 For example, Johnson and Cook (1985) investigated the fracture of copper and steels and  
 39 proposed a failure criterion describing a monotonic dependence of  $\varepsilon_F^p$  on  $\chi$  in the  $[0;1.2]$   $\chi$ -  
 40 range (as well as on strain rate and temperature). Bao and Wierzbicki (2004) evidenced a non-  
 41 monotonic dependence of  $\varepsilon_F^p$  on  $\chi$  in the  $[-1/3;1]$   $\chi$ -range for AA2024-T351 aluminum alloy,  
 42 and accordingly Bao and Wierzbicki (2005) proposed a 2-regime based dependence of  $\varepsilon_F^p$  on  
 43  $\chi$ . In the same time, Wierzbicki and Bao (2005) reported a criterion accounting explicitly for  
 44  $\chi$  and  $L$  in the expression of  $\varepsilon_F^p$ . For the same aluminum alloy for which they obtain different  
 45 fracture conditions in the  $[0;1]$   $\chi$ -range, i.e. a monotonic decrease of ductility, Papasidero et  
 46 al. (2015) identified a  $\theta$ - and loading direction-dependent  $\sigma_F$  (and further  $\varepsilon_F^p$  for a given strain  
 47 hardening law) derived from a Mohr-Coulomb like fracture initiation surface. Based on tests  
 48 carried out on AA7108-T6 aluminum alloy in the  $[0;2]$   $\chi$ -range, Hooputra et al. (2004)  
 49 developed two competing  $\varepsilon_F^p$  related fracture criteria, involving  $\chi$  and  $\eta$ , respectively.  
 50 Fracture was assumed to occur as soon as one of the two criteria is met. They also documented  
 51 the influence of the loading direction and strain rate. In this context of anisotropic materials, Ha  
 52 et al. (2019) and Korkolis et al. (2010) underlined the importance of using an appropriate  
 53 anisotropic plasticity model, in order to get correct predictions of ductile failure.

54  
 55 A number of studies dedicated to the ductile fracture of Ti-6Al-4V titanium alloy can also be  
 56 found in the literature.

57 For example, Giglio et al(2012), and Allahverdzadeh et al. (2015) investigated the quasi-static  
58 fracture of bimodal Ti-6Al-4V for nil and positive  $\chi$  ranging from 0 to 1, based on combined  
59 tension or compression plus torsion tests on smooth or notched cylinders, as well as tensile tests  
60 on various types of flat samples. They attempted to describe the fracture locus using either Bao-  
61 Wierzbicki's model (2004), or a coupled plasticity-damage model. The errors on the predicted  
62 fracture displacement or rotation in both cases were often 20-50%, probably due to the use of a  
63 too basic constitutive model, i.e. not accounting for anisotropy. Later, Zhang et al. (2020) used  
64 the previous set of experimental data to identify a coupled plasticity-damage model, accounting  
65 for anisotropy, tension-compression asymmetry, as well as combined isotropic and kinematic  
66 hardening. The agreement between model predictions and experimental data for fracture was  
67 significantly improved. The same authors had developed an anisotropic thermo-elasto-  
68 viscoplastic model coupled with isotropic ductile damage, Zhang et al. (2018), but their  
69 application to Ti-6Al-4V was limited to yield surface predictions at various temperatures.  
70 Vilotic et al (2016) performed confined ring upsetting tests on Ti-6Al-4V. They observed that  
71 ductile fracture occurs even at  $\chi$  as low as -0.63, and reported a monotonic decrease of ductility  
72 as triaxiality rises from -0.63 to +0.43. Tancogne-Dejean et al (2016) investigated the quasi-  
73 static ductile fracture of additively-manufactured Ti-6Al-4V using various types of flat notched  
74 samples cut along various directions. They used a non-associated anisotropic Hill48 flow rule,  
75 which approximately fitted the load-displacement curves, and proposed a probabilistic Mohr-  
76 Coulomb like fracture initiation criterion and further  $\sigma_F$  leading to a  $\chi$  - and  $L$  -dependence of  
77  $\varepsilon_F^p$ . The experimental scatter in  $\varepsilon_F^p$  was mainly attributed to the microstructural heterogeneity  
78 of the material (presence of "macro-zones" with correlated grains orientations inside, inherited  
79 from former beta grains). Wang et al. (2019) documented the anisotropic and temperature-  
80 dependent ductile fracture of Ti-6Al-4V thin sheets between room temperature (RT) and 800°C,  
81 through tensile tests on smooth and notched samples, as well as shear tests, along various  
82 loading directions, plus Nakajima tests, at strain rates between 0.01 and 1s<sup>-1</sup>. They reported a  
83 nearly constant ductility up to 400°C, and a substantial rise beyond. The anisotropy of  $\varepsilon_F^p$  was  
84 quite significant for shear loading (with nearly a factor of two between the most and least ductile  
85 directions), but much less at  $\chi$  above 1/3 (less than 15% difference). They used the Barlat-  
86 Lian1989 anisotropic yield criterion with an associated flow rule, and an incremental damage  
87 accumulation model, plus a localization criterion to determine the fracture conditions, with  
88 some success, but also some discrepancies. Chen et al. (2018) investigated the ductile fracture  
89 of bimodal Ti-6Al-4V in compression between 2.5x10<sup>3</sup> and 10<sup>5</sup> s<sup>-1</sup> at room temperature or at  
90 100°C. They reported that  $\varepsilon_F^p$  increases with the strain rate, and is higher at 100°C than at room  
91 temperature. They proposed a coupled plasticity-damage model, but they did not investigate  
92 anisotropy effects nor the quasi-static regime. Huang et al. (2018) studied the effect of  $\chi$  and  
93 strain rate on the ductile fracture of bimodal Ti-6Al-4V at room temperature, 150°C, and 300°C  
94 through quasi-static and dynamic tensile tests on smooth or notched cylindrical specimens, and  
95 reported a monotonic decrease of  $\varepsilon_F^p$  as  $\chi$  rose from 0.45 to 1.05. Ductility increased with the  
96 temperature, and was higher at high strain rates than in the quasi-static regime. They used  
97 Johnson-Cook plasticity and damage models, see Johnson and Cook (1985), to reproduce the

98 data. They however did not investigate anisotropy effects. Gross and Ravi-Chandar (2016)  
99 performed simulations of tension and shear tests run in the framework of the 2014 Sandia  
100 Fracture Challenge at two different strain rates in various loading directions, using Hill48 flow  
101 rule, the Johnson-Cook model (1985) to capture strain rate and temperature effects, plus a  $\varepsilon_F^p$ ,  
102 assumed to decrease monotonically with  $\chi$  in the range -0.63 to 3, and independent of the  
103 loading direction. The agreement with experimental data was incomplete, which was attributed  
104 to an imperfect modelling of plastic flow. Pack and Roth (2016) analysed the same test series,  
105 using the Hill'48 anisotropic yield function and the Johnson-Cook strain rate hardening and  
106 temperature softening functions, as well, but fitted a strain rate dependent Mohr-Coulomb like  
107 fracture initiation criterion. They conclude that  $L$  cannot be neglected.

108

109 From this brief literature survey, it appears that ductile fracture of Ti-6Al-4V alloy is generally  
110 reproduced by decoupled approaches of viscoplasticity and damage rarely correlated with  
111 microscopic observations, and that in spite of a substantial number of studies, a complete picture  
112 of the combined effects of the loading direction, strain rate, stress state on ductile damage and  
113 fracture is still missing for Ti-6Al-4V titanium alloy. Since all these aspects are crucial for the  
114 design of impact-resistant aircraft engine fan blades, an extensive experimental study was run  
115 in the present work in a wide range of  $\chi$ , strain rate, and loading directions, as representative  
116 of possible accidental conditions for aircraft fan blades as possible.

117 Within an engineering-oriented approach, two questions had to be addressed. Is there any need  
118 for a coupled damage-plasticity model, or is a  $\varepsilon_F^p$ -based fracture criterion sufficient? And in  
119 the latter case, is it possible to express the fracture criterion as a function of the stress triaxiality  
120 ratio  $\chi$  or/and Lode parameter  $L$ , as usually done?

121 In an attempt to answer these questions, a large series of tests, either leading to fracture, or  
122 interrupted just before fracture were carried out, and followed by the microscopic observation  
123 of the damage mechanisms. In parallel, numerical analyses were conducted aiming at  
124 investigating the influence of the loading direction, stress state ( $\chi$  ranging from -0.5 to 1), and  
125 strain rate (from  $10^{-3}$  to  $1.5 \times 10^3 \text{s}^{-1}$ ) on the ductility of Ti-6Al-4V titanium alloy.

126

127 The paper is organised as follows. In Section 2 the experimental procedure, including the  
128 material mechanical response in terms of stress-strain and load-displacement curves, is briefly  
129 presented. The damage and fracture mechanisms are then described in details in Section 3.  
130 Section 4 gathers the numerical simulations of some representative loading cases for the  
131 extraction of local values unreachable by the experiments. In Section 5 the experimental and  
132 numerical results are combined to identify the fracture locus of the material. Concluding  
133 remarks are finally gathered in Section 6.

## 134 2 Experimental investigation

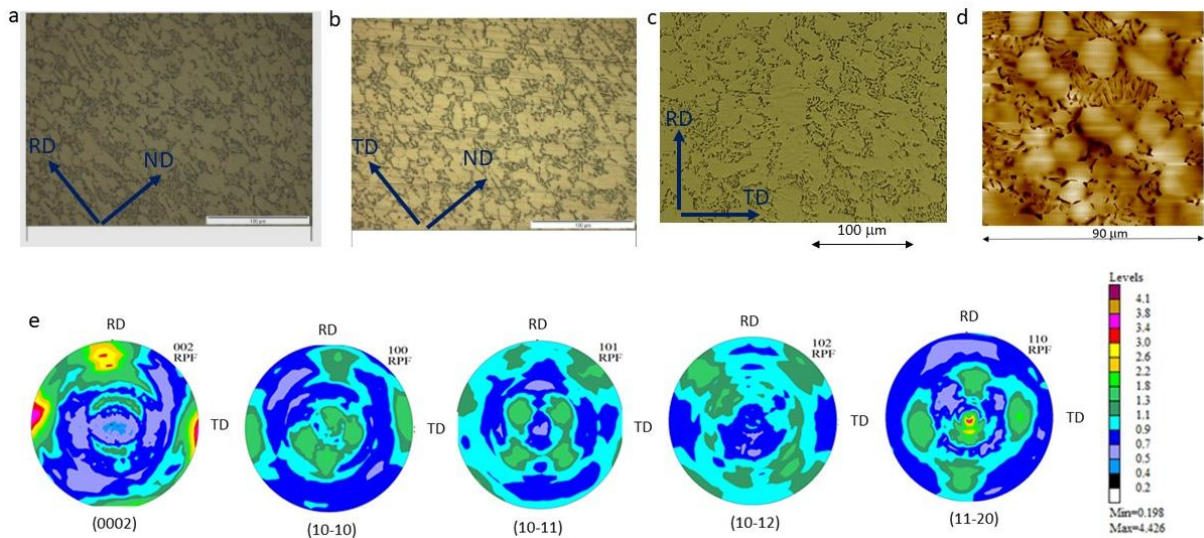
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### 136 2.1 Material under consideration and geometry of the specimens

137

138 The Ti-6Al-4V titanium alloy under consideration was provided in the form of a 16 mm-thick

139 cold-rolled sheet. Its microstructure is bimodal (see Figure 1), and the size of the equiaxed  
 140 primary  $\alpha$  phase nodules ranges from a few microns up to 30  $\mu\text{m}$ . Secondary  $\alpha$  laths, 5 to 20 $\mu\text{m}$   
 141 long and a few  $\mu\text{m}$  wide are visible within the partially transformed  $\beta$  phase (dissolved by the  
 142 chemical attack) on the atomic force microscopy (AFM) image (Figure 1d). The spatial  
 143 distribution of the  $\alpha$  and  $\beta$  phases does not seem completely random, since in many places, rows  
 144 of  $\alpha$  nodules parallel to the rolling direction (RD) or (in-plane) transverse direction (TD), and  
 145 perpendicular to (through-thickness) normal direction (ND) can be observed, in-between  
 146 elongated areas of partially transformed  $\beta$  phase (Figure 1a and b). Electron backscattering  
 147 diffraction (EBSD) mappings of grains orientations (not reproduced here, see Figure1 in [1])  
 148 reveal the existence of hundreds of  $\mu\text{m}$  wide "macro-zones" with a different orientation,  
 149 inherited from the orientation of the former  $\beta$  grains. However, even though the texture is  
 150 locally pronounced, the global texture measured by X-ray diffraction is actually moderate (the  
 151 maximum density of poles is at most 4.4 times that for a random texture), with the  $c$  axes of  
 152 HCP crystals mostly normal to ND, and clustered along RD or TD.  
 153



154  
 155 Figure 1: a-c) Optical images of the microstructure a) in the RD-ND plane, b) in the TD-ND  
 156 plane, c) in the RD-TD plane, d) AFM image in the RD-TD plane, and e) pole figures for the  
 157  $\alpha$  phase obtained by X-ray diffraction  
 158

159 Various types of specimens have been machined along four directions: the RD, TD and ND, as  
 160 well as a (in-plane) diagonal direction (DD) in the RD-TD plane, at 45° with respect to the  
 161 rolling direction. The specimens have been designed in order to investigate a wide range of  $\chi$   
 162 and  $L$ , viz.  $\chi \in [-1,1.2]$  and  $L \in [-1,1]$ . As a reminder, Table 1 reports the values of  $\chi$ ,  $\eta$   
 163 and  $L$  for a few loading cases.  
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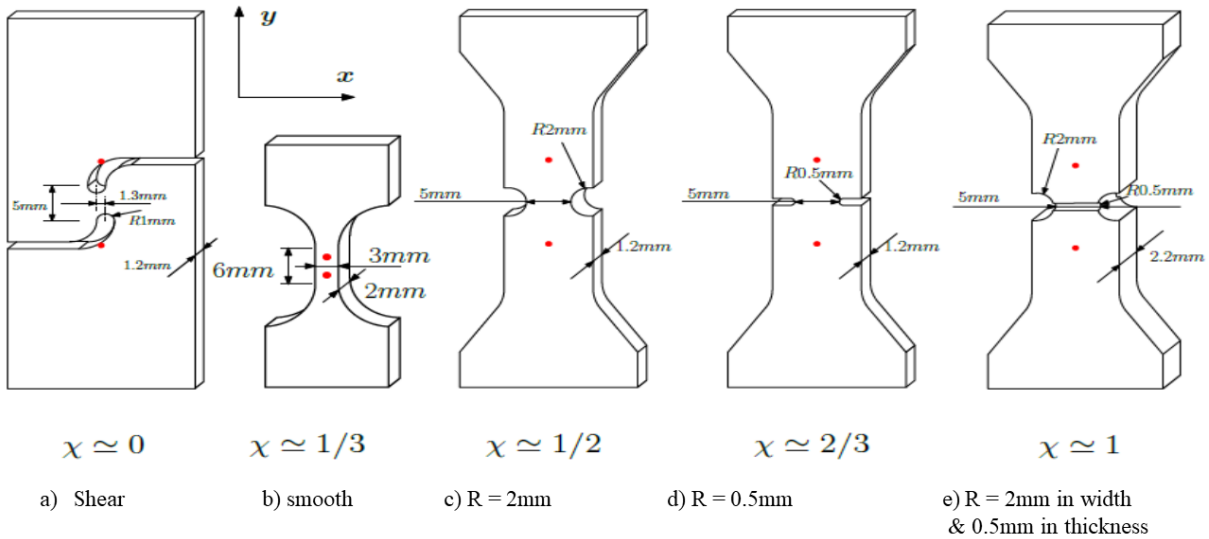
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**Table 1:** A few loading types and their corresponding values of stress triaxiality ratio  $\chi$ , shear stress ratio  $\eta$ , and Lode parameter,  $L$ .

	$\sigma_1$	$\sigma_2$	$\sigma_3$	$p$	$\tau_{\max}$	$q = \sigma_M$	$\chi$	$\eta$	$\chi/\eta$	$L$
Equi-biaxial compression	0	$-\sigma$	$-\sigma$	$2\sigma/3$	$\sigma/2$	$\sigma$	$-2/3$	$1/2$	$-4/3$	$-1$
Compression	0	0	$-\sigma$	$\sigma/3$	$\sigma/2$	$\sigma$	$-1/3$	$1/2$	$-2/3$	$1$
Shear	$\tau$	0	$-\tau$	0	$\tau$	$\sqrt{3}\tau$	0	$1/\sqrt{3}$	0	0
Tension	$\sigma$	0	0	$-\sigma/3$	$\sigma/2$	$\sigma$	$1/3$	$1/2$	$2/3$	$-1$
Equi-biaxial tension	$\sigma$	$\sigma$	0	$-2\sigma/3$	$\sigma/2$	$\sigma$	$2/3$	$1/2$	$4/3$	$1$

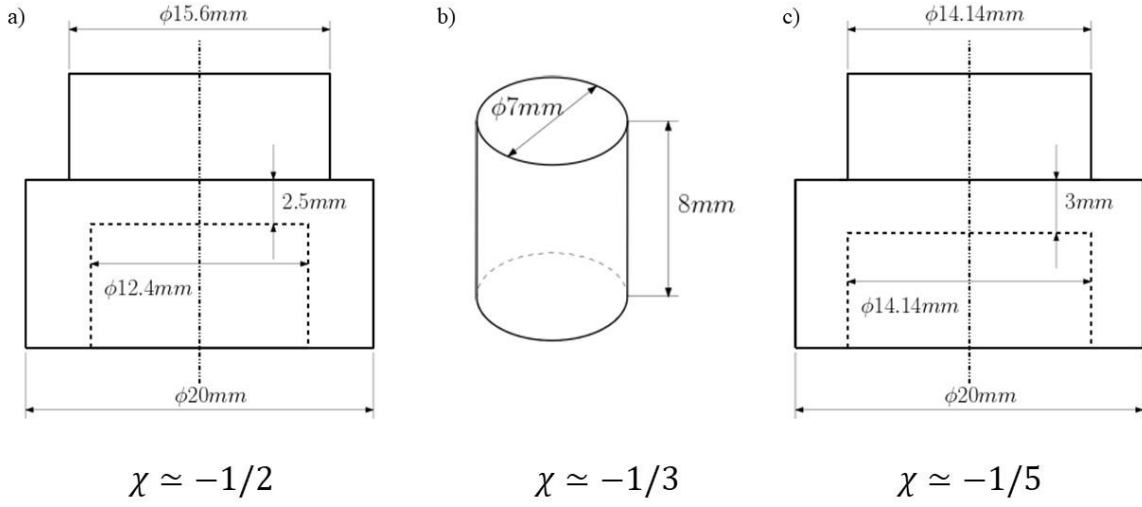
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Figure 2 shows the geometries of the specimens leading to nil and positive  $\chi$ , with the tension-induced shear specimen inspired from the work of Roth and Mohr(2018), and Figure 3 those leading to negative  $\chi$  inspired from Couque (2003) and Meyers et al (2001). Shorter smooth tensile specimens were used for the normal direction, along which the length is limited by the plate thickness. The cross section remained the same, but the gauge length was reduced to 4mm. The average values of  $\chi$  indicated in these figures were obtained from preliminary finite element simulations using the constitutive model detailed in Ruiz de Sotro et al (2020-a) and outlined in Appendix. The thickness of the flat specimens is not very large compared to the size of the “macro-zones”, but it is comparable to the thickness of the leading edge of the fan blades.



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Figure 2: Flat specimen geometries used to cover the triaxiality range [0,1]. The red dots indicate the gage length of the extensometer.



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188  
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Figure 3: Axisymmetric specimen geometries used to cover the stress triaxiality range  $\chi \in [-1/2, 0]$ . Hat-shaped specimens have been designed following Couque (2003) and Meyers et al. (2001)

192 Tension, compression and shear tests were carried out at room temperature and at low strain  
193 rate (typically ranging between  $|\dot{\epsilon}| \approx 10^{-4} s^{-1}$  and  $|\dot{\epsilon}| \approx 10^{-1} s^{-1}$ ) using standard tension-  
194 compression testing machines, and at high strain rate (typically of the order of  $|\dot{\epsilon}| \approx$   
195  $1.5 \times 10^3 s^{-1}$ ) using split Hopkinson pressure bar (SHPB) set-ups. In order to assess  
196 repeatability, two to four test repetitions were carried out for most tests. In the case of  
197 compression tests, a clip-on extensometer was mounted on the rigid plates compressing the  
198 sample, and grease was applied as lubricant on the top and bottom sides in order to minimize  
199 the barreling effect. For dynamic tension tests, the load-inversion device developed by Dunand  
200 et al. (2013) and modified by Roth et al. (2015) was used. In these tests, the strain field in the  
201 sample was obtained by digital image correlation (DIC) using Vic-2D software, based on  
202 images captured by a Phantom v7.3 high speed camera. A frame rate of  $10^5$  Hz with a resolution  
203 of  $304 \times 64$  pixel<sup>2</sup> was employed to monitor a  $10 \times 3$  mm<sup>2</sup> area covered with speckle painting. The  
204 global strain in the gage length was measured using a virtual extensometer based on the distance  
205 between two dots on the specimen, see Figures 2-3. Table 2 summarizes the test conditions.

206  
207

**Table 2:** Loading conditions

Specimen type	Loading direction	Strain or displacement rate
Shear	RD, TD, DD	$1.6 \times 10^{-3}$ mm/s
Smooth tension	RD, TD, DD ND	$10^{-3} s^{-1}$ , $10^{-2} s^{-1}$ , $3 \times 10^2 s^{-1}$ , $10^3 s^{-1}$ , $1.5 \times 10^3 s^{-1}$ $10^{-3} s^{-1}$ , $10^{-2} s^{-1}$
Simply notched 2mm	RD, TD, DD	$1.6 \times 10^{-3}$ mm/s
Simply notched 0.5mm	RD, TD, DD	$1.6 \times 10^{-3}$ mm/s
Doubly notched 0.5mm	RD, TD, DD	$1.6 \times 10^{-3}$ mm/s
Smooth compression	RD, TD, DD, ND	$10^{-3} s^{-1}$ , $10^{-2} s^{-1}$ , $3 \times 10^2 s^{-1}$ , $10^3 s^{-1}$
Meyer compression	RD, TD, ND	$10^{-3} s^{-1}$ , $10^3 s^{-1}$



Couque compression	RD, TD, ND	$10^{-3} \text{ s}^{-1}, 10^3 \text{ s}^{-1}$
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208

209 2.2 Mechanical characterization

210

211 In this section, the stress-strain and load-displacement curves obtained under tension,  
 212 compression and shear loading on smooth and notched specimens at various strain rates are  
 213 reported. For confidentiality reason, stress and load values are normalised with respect to the  
 214 yield stress (or load) along RD at room temperature, at a strain rate of  $10^{-3} \text{ s}^{-1}$ . For notched and  
 215 shear specimens, the term “apparent strain” refers to the relative displacement measured by the  
 216 virtual extensometer divided by its gage length (distance between the two reference dots).

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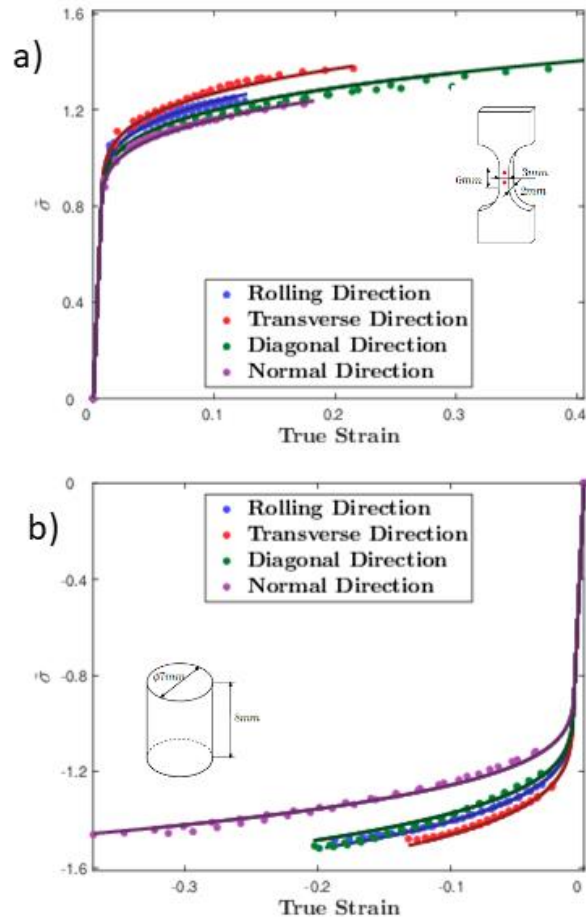
218 First, the influence of the loading direction and strain rate on the flow curves under uniaxial  
 219 compression or tension is depicted in Figures 4 and 5. The end point of each curve corresponds  
 220 to fracture. The curves obtained by simulation using the constitutive model outlined in  
 221 Appendix are superimposed in Figure 4. According to Figure 4, the material exhibits strain  
 222 hardening, and a higher yield stress along TD (along which most of the c axes, - known as a  
 223 “hard direction” of the  $\alpha$  phase- cluster), followed by RD (along which c axes also cluster, but  
 224 to a lesser extent), DD and finally a lower yield stress along ND (normal to most c axes). It also  
 225 shows a significant tension-compression asymmetry.

226 According to Figure 5, the material also exhibits strain rate hardening, as expected for a titanium  
 227 alloy. This strain rate dependence for RD was also observed for the other loading directions.

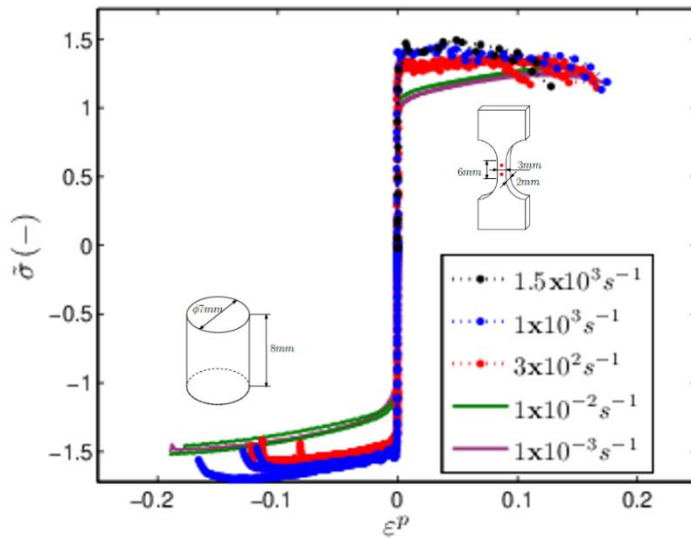
228 All these specificities, as well as a pronounced kinematic hardening (not documented here)  
 229 were captured by the constitutive model detailed in Ruiz de Sotro et al. (2020-a) and more  
 230 briefly recalled in Appendix.

231

232



233  
 234 Figure 4: Normalized stress-strain curves along various loading directions at  $10^{-3}s^{-1}$  in a)  
 235 tension and b) compression. The solid line corresponds to simulations using the constitutive  
 236 model described in Appendix  
 237



238  
 239 Figure 5: Normalized stress-strain curves for tension and compression along RD at various  
 240 strain rates

241 The macroscopic fracture strain and final necking-induced variations in width  $w$ , thickness  $t$ ,

242 and cross-section area  $A$ , of the smooth specimens broken under tension along RD, TD, ND or  
 243 DD at different loading rates are reported in Table 3. They are expressed as  
 244

$$\begin{cases} \eta_w = \frac{w_f - w_o}{w_o} \\ \eta_t = \frac{t_f - t_o}{t_o} \\ \eta_A = \frac{A_f - A_o}{A_o} \end{cases} \quad \text{Eq. 1}$$

245 where the subscripts  $o$  and  $f$  stand for initial and final (at fracture). The variation of the axial  
 246 fracture strain with the loading direction and strain rate is also plotted in Figure 6, without any  
 247 averaging of duplicate test results.  
 248

249 **Table 3:** Fracture strain and necking of smooth tensile specimens. The nominal dimensions  
 250 were used as a reference for the quasi-static tests. Since the tests have been repeated, the values  
 251 are averages over the number of broken samples.  
 252

Strain Rate		$T = 25^\circ C$		
		$10^{-3} s^{-1}$	$350 s^{-1}$	$1500 s^{-1}$
Fracture Strain $\varepsilon_N^F$	RD	16.9%	16.2%	13.4%
	DD	44.2%	20.7%	14.1%
	TD	21.2%	15.0%	18.8%
	ND	16.7%		
$\eta_w = \frac{\omega_f - \omega_o}{\omega_o}$	RD	16.1%	11.0%	12.4%
	DD	20.7%	17.0%	18.4%
	TD	16.1%	8.6%	19.0%
Thickness $\eta_t = \frac{t_f - t_o}{t_o}$	RD	23.1%	17.1%	26.6%
	DD	29.7%	18.5%	19.9%
	TD	24.5%	18.8%	33.8%
Cross-section $\eta_A = \frac{A_f - A_o}{A_o}$	RD	35.4%	26.2%	35.8%
	DD	44.3%	32.4%	34.6%
	TD	36.7%	25.8%	46.3%

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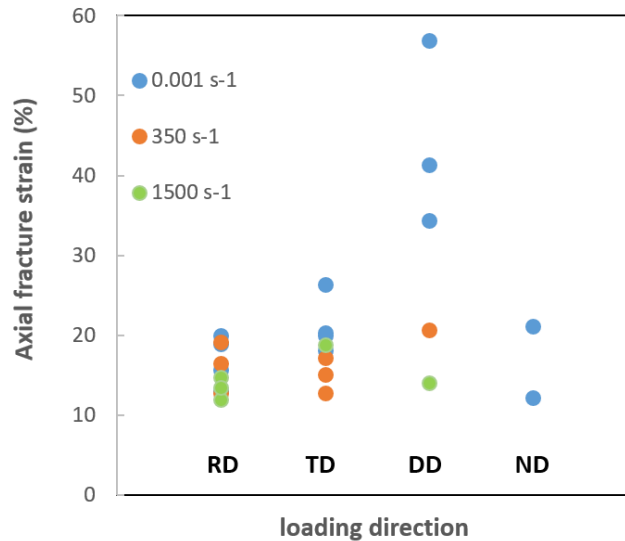


Figure 6: Variation of the axial tensile fracture strain with the loading direction and strain rate

The scatter from one sample to the other for the same loading conditions, as well as the different observed trends, depending on the criterion used to assess ductility (axial fracture strain, reduction in width, thickness or cross section) make the analysis of the anisotropy in terms of resistance to fracture of the material somewhat tricky. The scatter in experimental data is attributed to the presence of the aforementioned “macro-zones”, as the dimensions of these zones approaches the thickness of the tested specimens, which is itself comparable to that of the leading edge of the fan blades.

The only clear characteristic is a nearly twice higher ductility along DD than along any other direction. This might be because DD is more favourable to prismatic glide than RD and TD (along which (10-10) directions tend to cluster) and has the possibility to activate basal slip to accommodate the plastic strain, contrary to all three other directions (since the c axes are mostly along RD and TD, see Figure 1e). Wang et al. (2019) also underlined the highest ductility of a thin Ti-6Al-4V sheet along DD.

Available data do not evidence any significant difference in tensile ductility between RD and TD, which is consistent with the quasi symmetry of the pole figures (Figure 1e), and the absence of a morphological difference of the microstructure between those two directions.

The limited amount of tensile data along ND (due to the impossibility, with the available experimental set-up, to perform high strain rate tests on such small specimens) does not allow any conclusion. The thickness reduction for specimens pulled along RD, TD and DD is more pronounced than their width reduction, which is consistent with the easiest plastic flow along ND (see Figure 4).

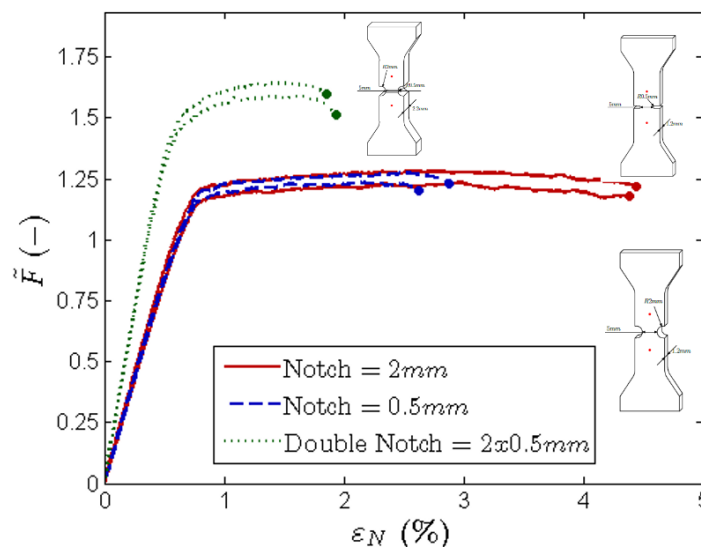
Contrary to what is observed on those smooth tensile specimens, for which the ductility along ND does not seem very different than along RD and TD, smooth cylindrical specimens compressed along ND failed at a much higher strain (38%) than along RD (25%), TD (16%) or even DD (28%). This is consistent with the results of Coghe et al. (2012) who also reported a higher compressive ductility of Ti-6Al-4V along a direction normal to most c axes, due to the

284 activation of  $\{10\bar{1}2\} \langle -1011 \rangle$  “tensile twins” (a misleading term which means that these twins  
 285 induce an extension along the c axis). Therefore, Ti-6Al-4V not only exhibits an anisotropic  
 286 ductility, but this anisotropy is different in tension (maximum ductility along DD) and in  
 287 compression (maximum ductility along ND).

288  
 289 Regarding the effect of the loading rate, no simple rule can be drawn either, since based on the  
 290 reduction in width and thickness, ductility seems to drop at  $\dot{\epsilon} \approx 350s^{-1}$  but to rise again at  $\dot{\epsilon} \approx$   
 291  $1500s^{-1}$ , sometimes reaching higher values than in the quasi static regime. This rise cannot be  
 292 attributed to an enhancement of ductility by self-heating, since the temperature has been  
 293 estimated to rise by at most  $140^{\circ}C$  at  $1.5 \times 10^3 s^{-1}$  Ruiz de Sotro (2020-b), and additional tensile  
 294 tests run at  $170^{\circ}C$  yielded approximately the same ductility as at room temperature. The rise  
 295 might perhaps be due to micro-inertia effects that delay microvoids coalescence, as suggested  
 296 by Jacques et al.(2012) . It might also result from an increasing accommodation of plastic strain  
 297 by mechanical twinning, shown by Coghe et al. (2012) to become twice more active at  $10^3 s^{-1}$   
 298 than at  $10^{-3} s^{-1}$  in Ti-6Al-4V. Note that Huang et al. (2018) reported a higher ductility for Ti-  
 299 6Al-4V at  $10^3 s^{-1}$  than at  $10^{-3} s^{-1}$ .

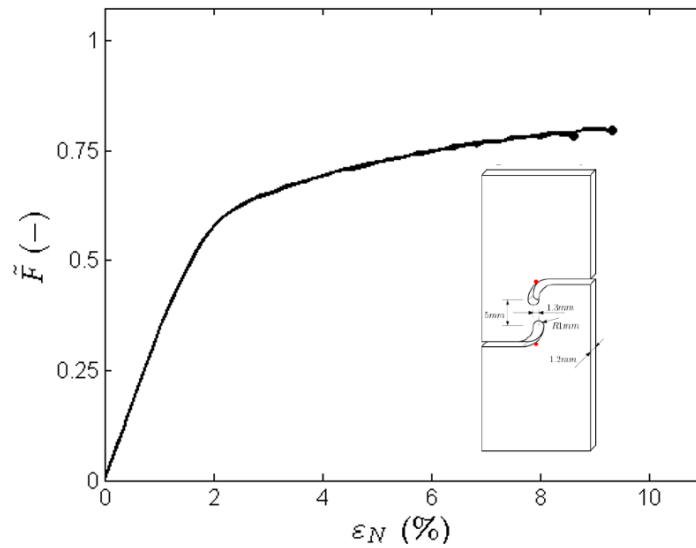
300 By order of importance of its effect on ductility, it seems that the loading direction with respect  
 301 to the cold-rolled sheet (thickness as opposed to width) comes first, then the in-plane loading  
 302 direction (RD, DD, TD), followed by the loading rate, for which no rule can be drawn.

303  
 304 Figure 7 shows the normalized force vs. apparent strain for the flat, notched tension specimens  
 305 pulled along TD until fracture. As expected, the resistance and apparent strain at fracture are  
 306 highly dependent on the notch radii, and resulting  $\chi$  . The highest force and lowest nominal  
 307 strain at fracture are observed for the doubly notched specimen ( $\chi \approx 1$ ), while both simply  
 308 notched specimens ( $\chi \approx 1/2, 2/3$ ) exhibit the same peak load, but a lower fracture strain for  
 309 the smallest notch radius. Similar effects were observed for the other loading directions.



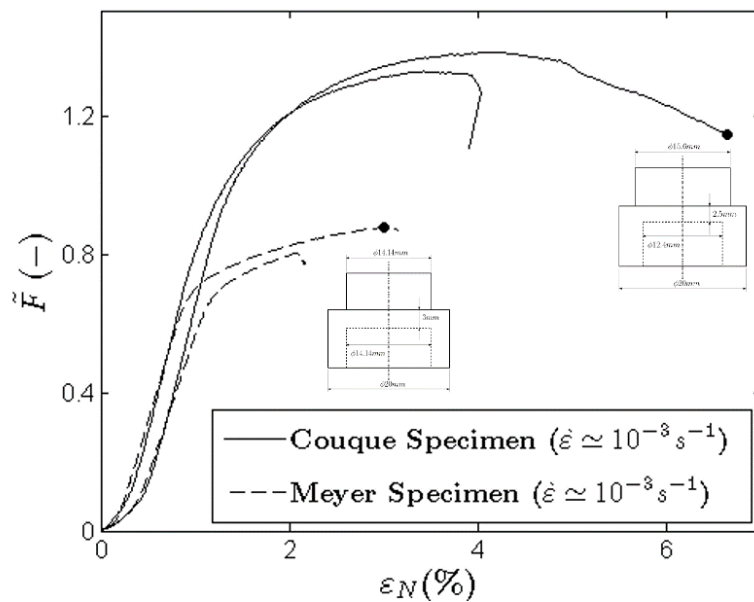
310  
 311 Figure 7: Normalized load versus apparent strain for tensile tests along TD on notched  
 312 specimens. Displacement rate:  $1.6 \times 10^{-3} mms^{-1}$ . Fracture is denoted by a circle  
 313

314 Figure 8 shows the load-apparent strain curves for two tension-induced shear specimens pulled  
 315 until fracture along TD. Strain hardening, associated with a large strain at fracture can be  
 316 noticed (and the same observation was made in the other loading directions).



317  
 318 Figure 8: Normalized load versus apparent strain for shear specimens pulled along TD.  
 319 Displacement rate:  $1.6 \times 10^{-3} \text{mms}^{-1}$ . Fracture is denoted by a circle

320  
 321 Finally, Figure 9 shows the normalized force vs. apparent strain curves for Couque ( $\chi \simeq -1/2$ )  
 322 and Meyers ( $\chi \simeq -1/5$ ) type hat-shaped specimens. For each geometry, one test was carried  
 323 out until fracture and the other test was interrupted for further microscopic observations. As  
 324 expected, the combined effects of shear and compression on Couque's sample produce a higher  
 325 load (higher pressure), as compared to Meyers's geometry (lower pressure). As the two  
 326 geometries involve very different stress and strain states, the corresponding apparent axial strain  
 327 values cannot be directly compared.



328  
 329 Figure 9: Normalized load versus apparent strain for hat-shaped specimens compressed along

ND. Displacement rate:  $8 \times 10^{-3} \text{ mms}^{-1}$ . Fracture is denoted by a circle

### 2.3 Damage and fracture characterization

Observations of fractured samples, as well as observations after interrupted tests, were carried out using scanning electron microscopy (SEM). For that purpose, mechanical polishing with sandpaper, followed by electro-polishing (using Struers A3 solution and a tension of 20V) or ion polishing was thus performed on longitudinal sections of the samples. The results are reported with respect to  $\chi$ .

#### Positive stress triaxiality $\chi$

Figure 10 shows SEM observations of smooth tension specimens ( $\chi \approx 1/3$ ), see Figure 5 for the corresponding stress-strain curve. Voids are found near the fracture surface. They appear to nucleate mainly on interfaces between  $\alpha$  and  $\beta$  phases, although a few very small cavities are found within primary  $\alpha$  grains. This is consistent with the observations made by Helbert et al(1996). The loading rate does not seem to have any direct effect on void nucleation.

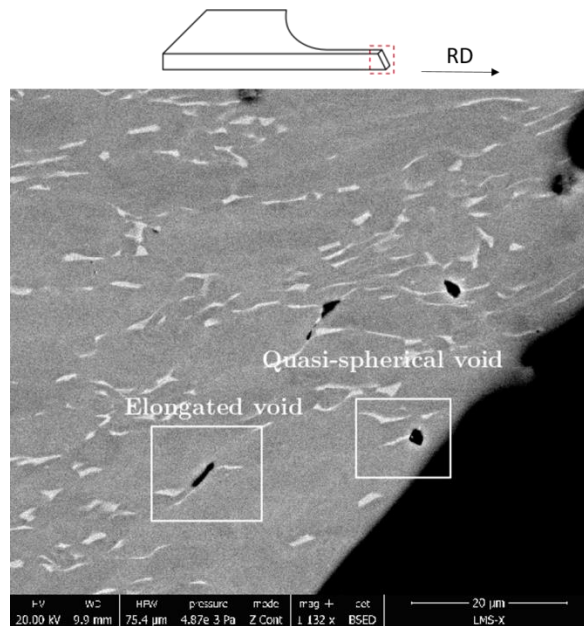
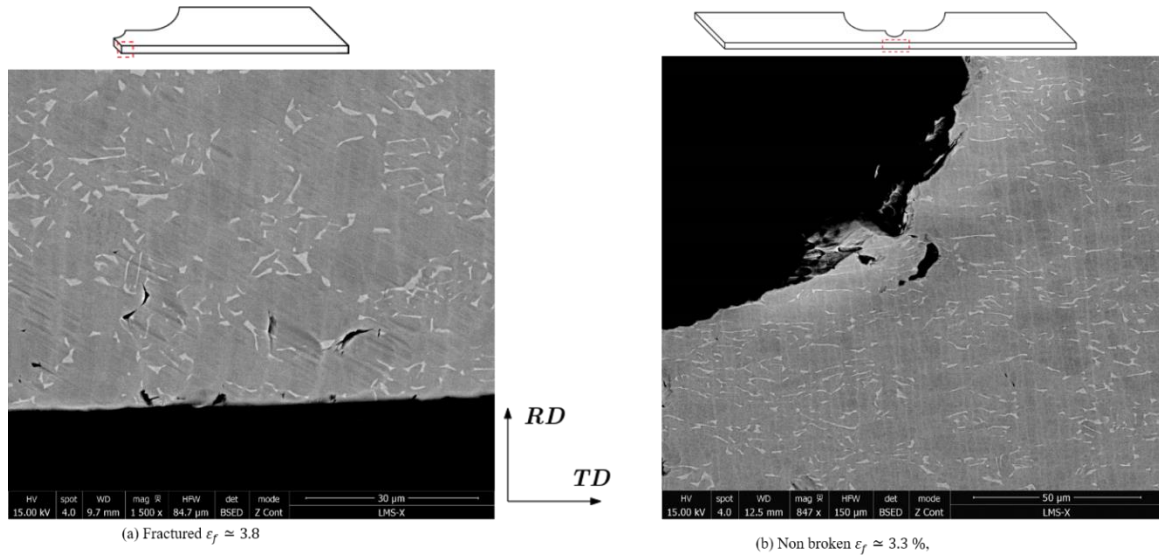


Figure 10: SEM observation of microvoids on a longitudinal section of a smooth tensile specimen fractured at  $350 \text{ s}^{-1}$  along RD

The same type of observations was carried out on notched TD specimens ( $\chi \approx 1/2, 2/3, 1$ ). As in the smooth specimens, some cavities are found near the fracture surface (see Figure 11a for  $\chi \approx 1/2$ ), see Figure 7 for the corresponding stress-strain curve. However, these cavities become scarcer away from the fracture surface, in spite of the considerably higher stress triaxiality ratio.

In an attempt to determine at which stage the cavities were formed, a test on a notched specimen was interrupted just before fracture ( $\epsilon \approx 3.3\%$ ) and a longitudinal section was observed (see

359 Figure 11b). Some cavities were found near the free surfaces, indicating that, although it  
 360 remains limited, diffuse damage took place before fracture, under high stress triaxiality ratios.  
 361



362  
 363 Figure 11: SEM observations of longitudinal sections of notched tension TD specimens  
 364 ( $R=2\text{mm}$ ,  $\chi \approx 1/2$ ,  $\dot{\epsilon} = 10^{-3} \text{s}^{-1}$ ).  
 365

366 Figure 12 presents macroscopic views of the broken smooth tension specimens ( $\chi \approx 1/3$ )  
 367 loaded along RD, TD and DD directions at low strain rate, at room temperature. Both the front  
 368 and side views are shown. For the three cases, the fracture surface is inclined into the thickness.  
 369 According to the side view, for both the RD and TD specimens the fracture surface is rather  
 370 smooth (a bifurcation is however visible for TD) and is inclined at  $45^\circ$ , while for the DD  
 371 specimen the fracture surface has a parabolic shape with two branches inclined at about  $\pm 45^\circ$ .  
 372 Shear strain localization, thus does not seem to follow the rows of  $\alpha$  nodules perpendicular to  
 373 ND mentioned above.  
 374



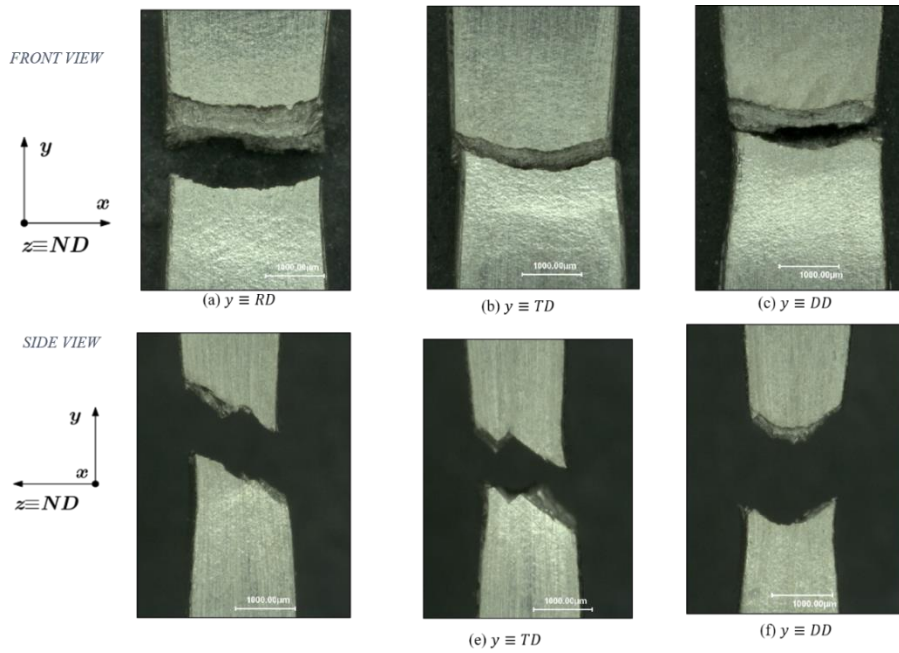


Figure 12: Broken smooth tensile specimens ( $\chi \approx 1/3$ ).  $\dot{\epsilon} = 10^{-3} \text{s}^{-1}$

Figure 13 shows the fracture surface of a notched TD specimen (notch radius 2mm,  $\chi \approx 1/2$ ). The surface is very rough showing heterogeneities of all sizes. A closer look reveals extremely dense clusters of more or less equiaxed dimples. These features suggest the nucleation, growth, and coalescence of voids favoured by a high hydrostatic tension. Furthermore, although no obvious signs of the fracture initiation site are found, it is speculated to be close to the center (expected maximum stress triaxiality) considering the large, uneven features around the middle, and the absence of directional features near the edges.

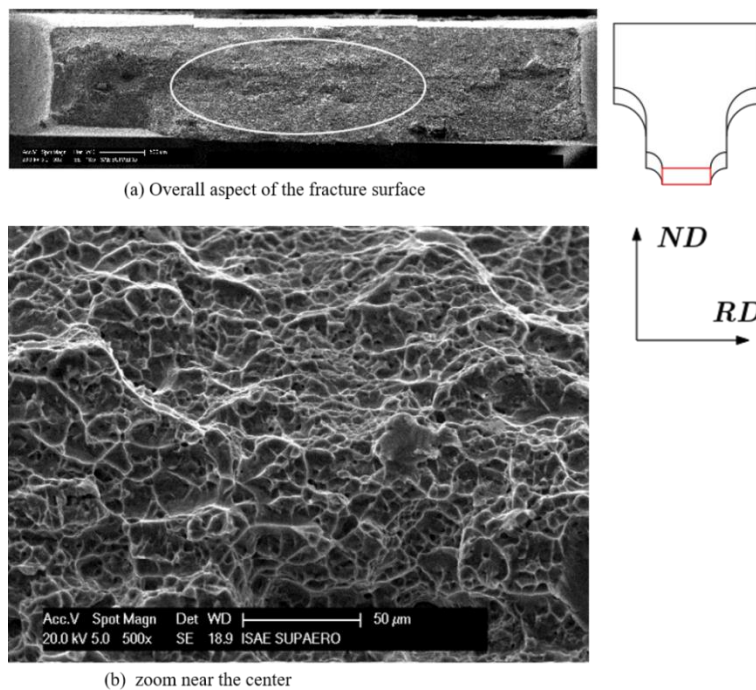


Figure 13: a) SEM micrograph of a notched TD specimen ( $R=2 \text{ mm}$ ,  $\chi \approx 1/2$ ).  $\epsilon_f \approx 3.8 \%$ ,  $10^{-3}$

388  $3s^{-1}$ . White ellipse: likely place of crack initiation, b) Zoom near the center

389

390 A similar observation was performed for the notched tension specimen of radius 0.5mm notch  
391 radius,  $\chi \approx 2/3$ , (see Figure 14). For this TD sample, the aspect of the surface seems  
392 comparatively more uniform than for the larger notch radius. The same type of equiaxed  
393 dimples is retrieved. In this case, fracture seems to have initiated from the notch roots, where  
394 the surface roughness is significantly larger. The reason behind this might be a larger plastic  
395 strain in this area, combined with a sufficiently high stress triaxiality ratio. The maximum  $\chi$  is  
396 speculated to occur away from the center-line. This hypothesis will be confirmed by numerical  
397 simulations presented in the following Section.

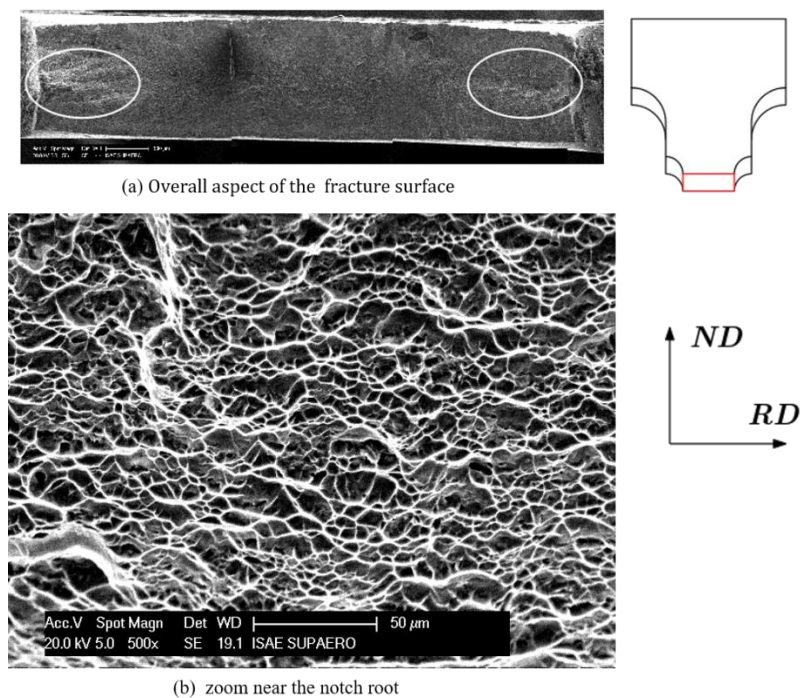
398

399 Nil stress triaxiality  $\chi$

400

401 A substantial distribution of cavities, or even micro-cracks, was observed in a tension-induced  
402 shear TD specimen ( $\chi \approx 0$ ) after an interrupted test (see Figure 15). This was somewhat  
403 unexpected, considering the rise of the critical equivalent plastic strain for cavity nucleation in  
404 Titanium alloys reported at low hydrostatic pressure by Helbert et al (1996).

405



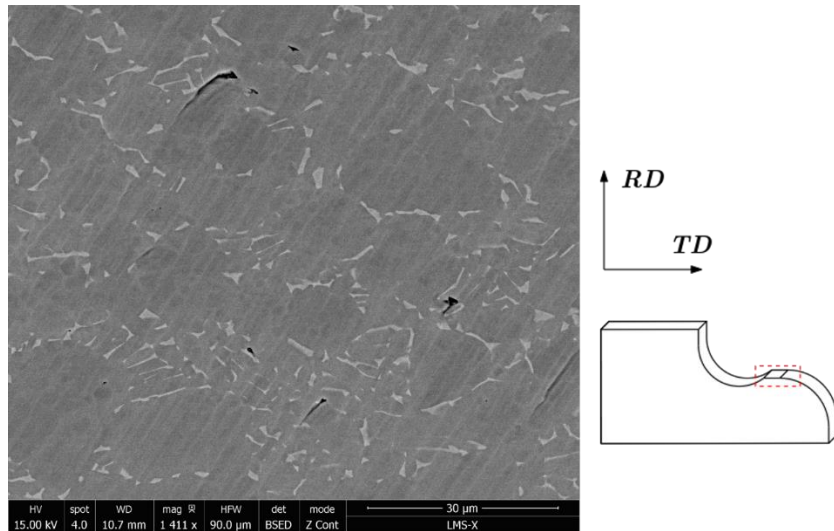
406

407 Figure 14: SEM micrograph of a notched TD specimen ( $R=0.5\text{mm}$ ,  $\chi \approx 2/3$ ).  $\dot{\epsilon}f \approx 2.5\%$ ,  $\dot{\epsilon} = 10^{-3}\text{s}^{-1}$ . White ellipses: likely places of crack initiation. b) Zoom near the notch root

408

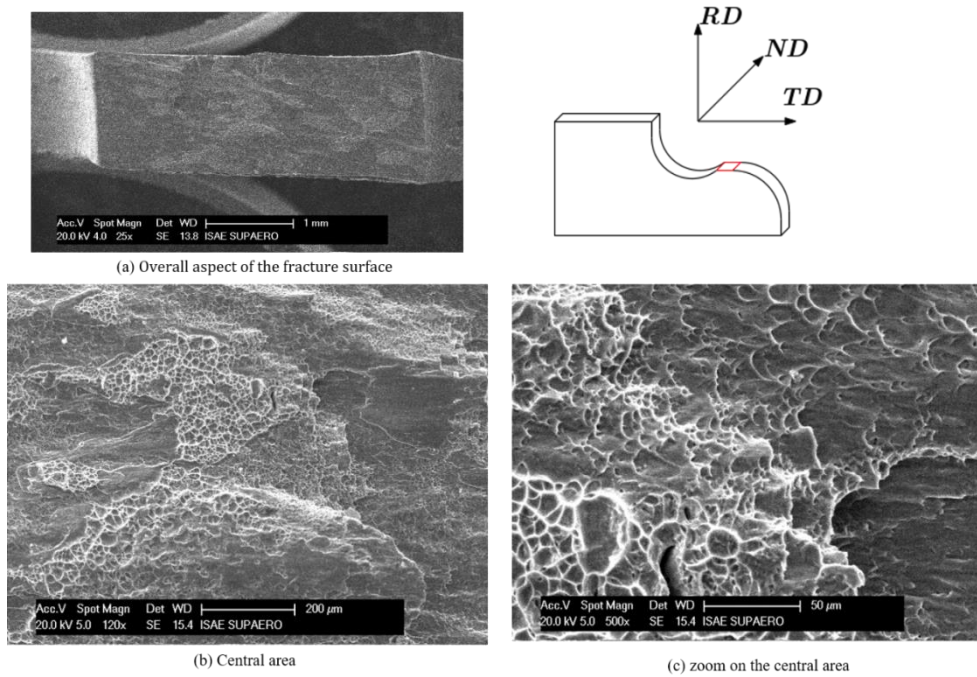
409

410



411  
 412 Figure 15: SEM observation of a longitudinal section of a shear TD specimen ( $\chi \approx 0$ ) after a  
 413 test interrupted at  $\varepsilon \approx 9\%$ ,  $\dot{\varepsilon} = 10^{-3} \text{s}^{-1}$ .  
 414

415 Figure 16 shows the smooth fracture surface of such a TD specimen, normal to the free surface  
 416 and parallel to the tensile axis. It looks heterogeneous, and exhibits clusters of elongated  
 417 dimples surrounded by smooth, mated zones.  
 418



419  
 420 Figure 16: SEM observations of a broken shear TD specimen ( $\chi \approx 0$ ).  $\varepsilon_f \approx 9\%$ ,  $10^{-3} \text{s}^{-1}$ .  
 421

422 Negative stress triaxiality  $\chi$

423

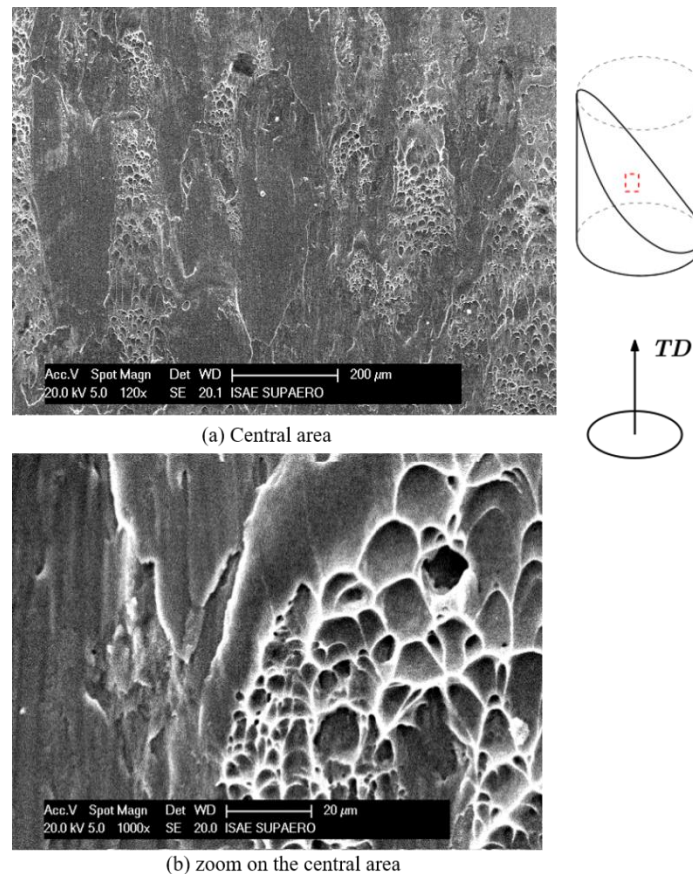
424 The fracture of smooth compression specimens ( $\chi \approx -1/3$ ) is sudden, precocious and localized



425 in a band inclined by  $45^\circ$  with respect to the compression axis. Figure 17 shows a SEM  
 426 observation of such a fracture surface, which exhibits mated areas surrounding elongated  
 427 dimple clusters. These micrographs show a great resemblance with those presented in Lee and  
 428 Lin(1997).

429 Compression tests on hat-shaped specimens for which macro crack propagation had begun were  
 430 successfully stopped (see Figure 18 for  $\chi \simeq -1/2$  at low and high strain rates). The macro  
 431 cracks, initiated from all inner and outer corners, and propagated (quasi) symmetrically along  
 432 a seemingly straight path joining the two corners. However, when zooming on the crack tip, a  
 433 staircase-shaped trajectory appears, as the likely result of the coalescence of some diffuse  
 434 micro-cracks nucleated under shear. Such a crack shape had been observed in other Ti-6Al-4V  
 435 alloys loaded at high strain rate (see Longère and Dragon(2015), Dorothy and Longère (2019),  
 436 and Peirs et al. (2010).

437  
 438 In the Meyers's geometry ( $\chi \simeq -1/5$ ) deformed under shear loading, multiple cracks also  
 439 nucleated from the corners of the specimen, and – as in the other hat-shaped geometry – seemed  
 440 to grow as the result of the coalescence of micro-cracks generated along their path.  
 441



442  
 443 Figure 17: SEM observations of the fracture surface of a compression TD specimen ( $\chi \simeq -1/3$ ).  
 444  $\epsilon_f \simeq 17\%$ ,  $\dot{\epsilon} = 10^{-3} \text{s}^{-1}$ .  
 445

446 2.4 Synthesis of the observations

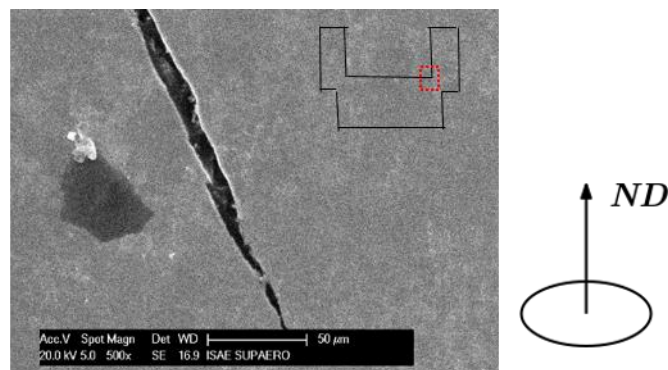
447

448 Comparing all the above presented micrographs, one can distinguish two different fracture  
 449 mechanisms described by many authors: (i) a shear-driven mechanism, under negative  $\chi$   
 450 (compression and shear loadings), and (ii) a cavity nucleation/growth and coalescence  
 451 mechanism, typical of positive  $\chi$ . This competition between shear localization and micro-  
 452 voiding must accordingly be accounted for in the formulation of a fracture criterion.

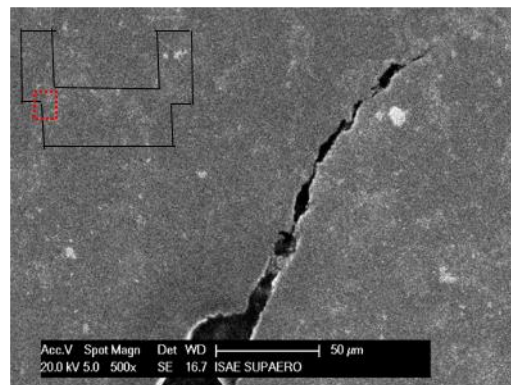
453 Damage, in the form of micro-cavities, nucleated mostly at  $\alpha/\beta$  interfaces, was found near the  
 454 fracture surfaces. The latter were often inclined by  $45^\circ$  into the thickness of smooth tension  
 455 specimens. Depending on the loading conditions, shear-driven strain localization and damage,  
 456 leading to slanted fracture with limited necking, or more diffuse, triaxiality-driven cavities  
 457 nucleation, growth and coalescence leading to fracture with more substantial necking were  
 458 observed. No clear conclusions can be drawn on the influence of strain rate, or even the loading  
 459 direction.

460 A significant ductility is evidenced, with only a weak micro-voiding induced damage activity  
 461 before failure. One can thus conclude that the coupling between micro-voiding induced damage  
 462 and elasticity/plasticity only plays a minor role, and may accordingly be neglected. In other  
 463 words, there seems to be no need for a coupled damage-plasticity model – see e.g. Longère et  
 464 al. (2012), Longère and Dragon (2016) and Hfaiedh et al. (2018) for situations wherein damage-  
 465 plasticity coupling is needed.

466



(a)  $\dot{u} = 0.5\text{mm/min}$



(b)  $\dot{u} = 5 \times 10^3 \text{mm/s}$

467

468

Figure 18: Couque’s hat-shaped ND specimens ( $\chi \approx -1/2$ ) after partial fracture. (a)  $\dot{u} =$

0.5mm/min , and (b)  $\dot{u} = 5 \times 10^3$  mm/s

469  
470

### 471 3 Numerical analyzes

472

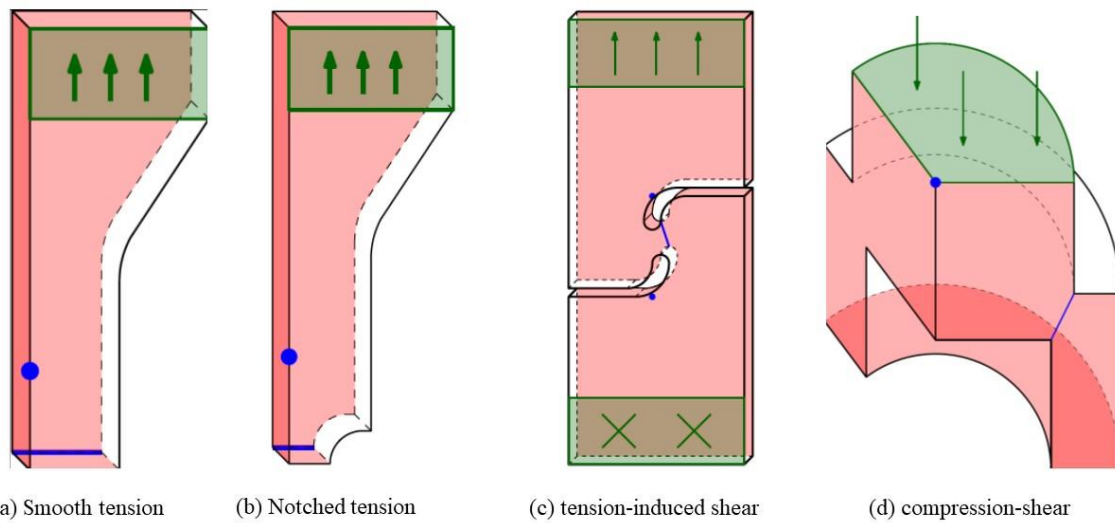
473 The constitutive model used to simulate the tests, detailed in Ruiz de Sotro et al. (2020-a) and  
474 outlined in Appendix, does not account for damage, but captures the anisotropic behavior, rate  
475 dependence, tension-compression asymmetry and isotropic/kinematic hardening of the  
476 material. It has been implemented as a user material subroutine in the explicit version of the  
477 commercial finite element code LS-DYNA LSTC (2018) and used for three-dimensional  
478 numerical simulations of the experimental configurations (using the mass scaling option for  
479 quasi-static tests), in order to estimate the local evolutions of the plastic strain  $\kappa$  (see Appendix  
480 for the definition), stress triaxiality ratio  $\chi$  and Lode parameter  $L$  in the critical area,  
481 unreachable by the experiment.

482

483 Selectively reduced (S/R) integration 8-node solid element (8 integration points with constant  
484 pressure throughout the element), also known as “elform 2” or bricks in LS-DYNA, are used.  
485 In order to check the mesh objectivity of the numerical results, various mesh sizes are  
486 considered.

487 While all loading cases reported in Section 2 have been numerically simulated, only four  
488 examples are detailed below for sake of brevity: (i) tension on flat smooth specimens, (ii)  
489 tension on flat notched specimen, and (iii) tension-induced shear loading on flat half-smiley  
490 specimen, and (iv) compression-shear on hat-shaped specimens.

491 Figure 19 shows the specimens’ sketches with the boundary conditions. Accounting for  
492 through-thickness, through-width and through-length symmetries of the flat, smooth and  
493 notched, tension specimens, only one eighth of the specimens is modelled with adequate  
494 boundary conditions. Accounting for through-thickness symmetry of the half-smiley specimen,  
495 only half of the specimen is meshed with adequate boundary conditions. Hat-shaped specimens  
496 exhibit a symmetry of revolution, but due to the plastic anisotropy of the material, axisymmetric  
497 simulations would not be correct. A quarter of the Couque and Meyer specimen are thus meshed  
498 with adequate boundary conditions. The controlled displacement (at constant velocity) is  
499 applied to the nodes of the top side while the bottom side is constrained in the vertical direction.

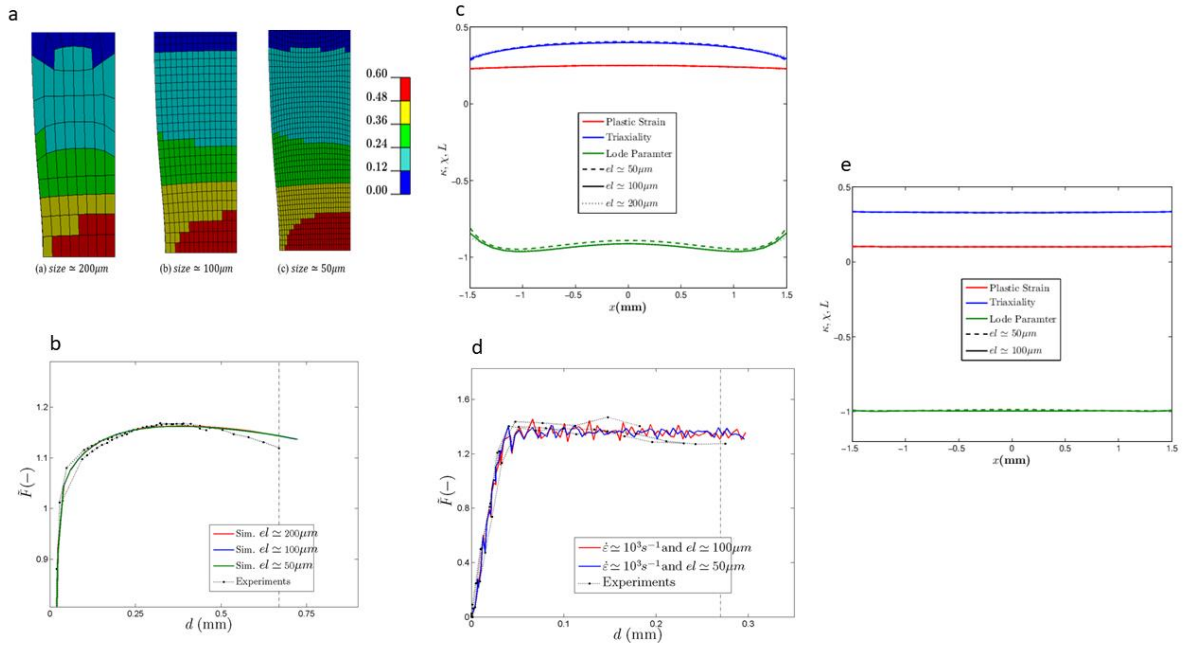


500 Figure 19: Geometries and boundary conditions for the numerical simulations. The green  
 501 areas represent the imposed displacement (or fixture) and the red zones the symmetry  
 502 conditions. The parts where internal variables are extracted are indicated by blue lines.  
 503

504  
 505 **3.1 Uniaxial tension**

506  
 507 The smooth tension specimens were modelled with three mesh sizes of 50, 100 and 200  $\mu\text{m}$  in  
 508 the gauge length. The axial plastic strain field in the three deformed meshes are compared in  
 509 Figure 20a at an advanced stage of deformation involving necking. The specimen grip part with  
 510 larger mesh sizes is not represented. In Figure 20b the experimental dimensionless force-  
 511 displacement curve for tension along RD at low strain rate is compared with the curves obtained  
 512 considering the three mesh sizes. It can be seen that the experimental curve is well captured,  
 513 and that there is no significant mesh dependence. This comment also applies to the other loading  
 514 directions, see Ruiz de Sotro et al. (2020).

515 Numerical simulations at high loading rate were conducted for mesh sizes of 50 and 100  $\mu\text{m}$  in  
 516 the gauge length. Figure 20d shows a comparison between the numerical and the experimental  
 517 results. Independently of the mesh size, the numerical curve is oscillatory. This simulation  
 518 artefact is due to the mechanical waves induced by the rapid loading that keep bouncing  
 519 between the boundaries. This undesirable effect may be fixed by modelling the whole specimen  
 520 and the input and output bars, so that the wave can be reflected/transferred from/outside the  
 521 specimen. The smoothing of such oscillations by averaging every three consecutive points  
 522 yields a curve close to the experimental one, with no significant mesh dependence.  
 523



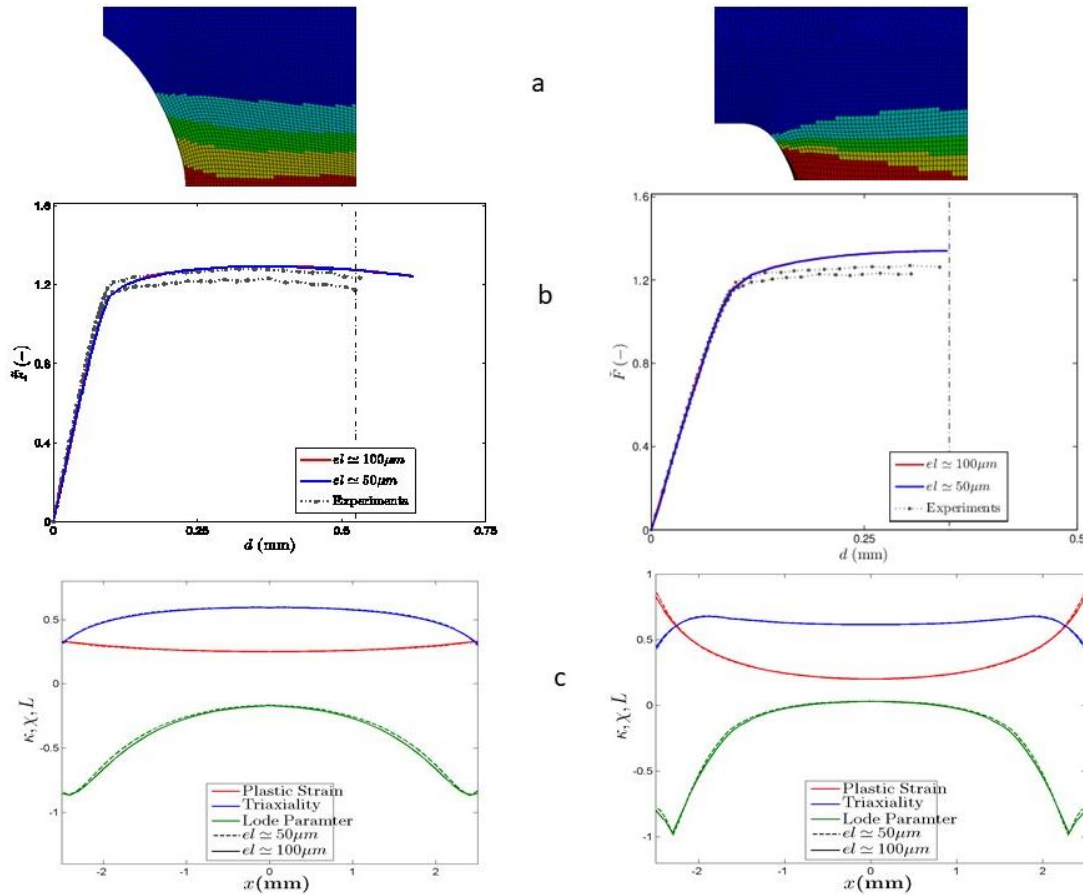
524  
 525 Figure 20: Simulation of the tensile tests along TD on smooth specimens, using three different  
 526 mesh sizes. a) comparison of the plastic strain fields at  $10^{-3}\text{s}^{-1}$ , b) normalized load-  
 527 displacement curves at  $10^{-3}\text{s}^{-1}$ , c) transverse profiles of plastic strain  $\kappa$ , stress triaxiality  $\chi$  and  
 528 Lode parameter  $L$  at mid-length at  $10^{-3}\text{s}^{-1}$ , the point where fracture is expected to initiate is  
 529 indicated by a vertical line, d) normalized load-displacement curves at  $10^3\text{s}^{-1}$ , e) transverse  
 530 profiles of plastic strain, Lode parameter and triaxiality at mid-length at  $10^3\text{s}^{-1}$   
 531

532 In addition, the plastic strain  $\kappa$ , stress triaxiality  $\chi$  and the Lode parameter  $L$  profiles computed  
 533 along the horizontal symmetry line of the specimen, at mid thickness converge for a mesh size  
 534 of  $100\ \mu\text{m}$  (see Figures 20d and e).

535  
 536 3.2. Tension on notched specimens  
 537

538 Figure 21 compares the experimental dimensionless load-displacement curves to the ones  
 539 obtained from simulations run with two different mesh sizes of  $50\ \mu\text{m}$  and  $100\ \mu\text{m}$ , for the notches  
 540 of  $R = 2\text{mm}$  ( $\chi \approx 1/2$ ) and  $R = 0.5\text{mm}$  ( $\chi \approx 2/3$ ). Within the displacement range covered  
 541 by the experimental curves, no effect of the element size is observed on the global scale.  
 542





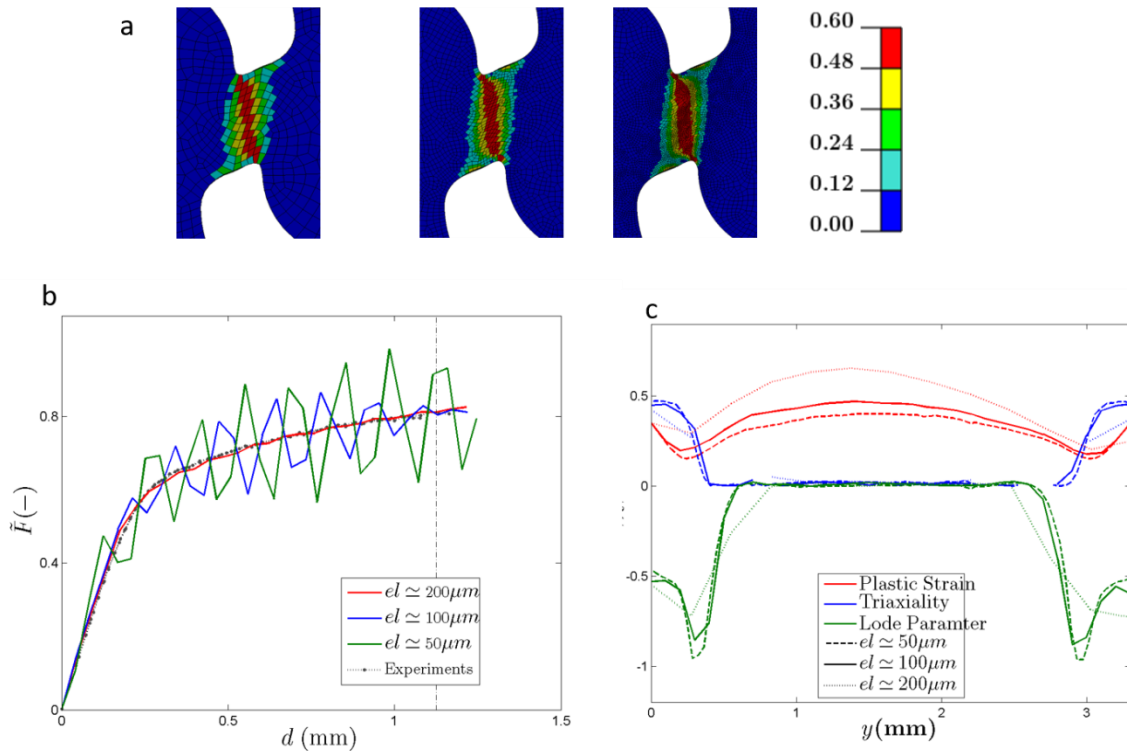
543  
 544 Figure 21: Simulation of the tensile tests along RD on simply notched specimens, using two  
 545 mesh sizes for  $R=2\text{mm}$  (left) and  $R=0.5\text{mm}$  (right) a) Plastic strain fields computed with the  
 546 fine mesh, b) normalized load-displacement curves, c) transverse profiles of plastic strain  $\kappa$ ,  
 547 stress triaxiality  $\chi$  and Lode parameter  $L$  at mid-length at the instant when fracture is expected  
 548 to initiate, indicated by a vertical line on Figure 21b

549  
 550 In Figure 21b, the plastic strain and the stress triaxiality ratio profiles along the width of simply  
 551 notched specimens are plotted from one notch root to the other at the instant when fracture is  
 552 expected to initiate, indicated by a vertical line on Figure 21b. Concerning the wider notch ( $R =$   
 553  $2\text{mm}$  and lower stress triaxiality  $\chi$ ), the maximum values of stress triaxiality are found at mid-  
 554 width, and the stress triaxiality  $\chi$  quickly decreases towards the notch roots. As for the plastic  
 555 strain, it is relatively uniform irrespective of the lateral position (especially at large strain).  
 556 A different trend is seen for the sharper notch ( $R = 0.5\text{mm}$  and higher  $\chi$ ). The stress triaxiality  
 557 remains quasi-constant over most of the width (the maximum values are found close to the  
 558 edges), whereas the plastic strain is substantially higher near the notch roots, where it is up to  
 559 three times larger than at the center.

560  
 561 

### 3.3 Tension-induced shear loading

562  
 563 The effect of the mesh size for the tension-induced shear specimen is shown in Figure 22 in  
 564 terms of strain field, computed dimensionless load-displacement curves, and profiles of the  
 565 plastic strain, Lode parameter and triaxiality along a vertical line joining two notch roots.



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Figure 22: Simulation of the tensile tests along RD on shear specimens, at  $1.6 \cdot 10^{-3} \text{ mms}^{-1}$ , using three mesh sizes a) Comparison of the plastic strain fields, b) normalized load-displacement curves, c) profiles of plastic strain  $\kappa$ , stress triaxiality  $\chi$  and Lode parameter  $L$  along a vertical line joining two notch roots. The instant when fracture is expected to initiate is indicated by a vertical line in Figure 22b

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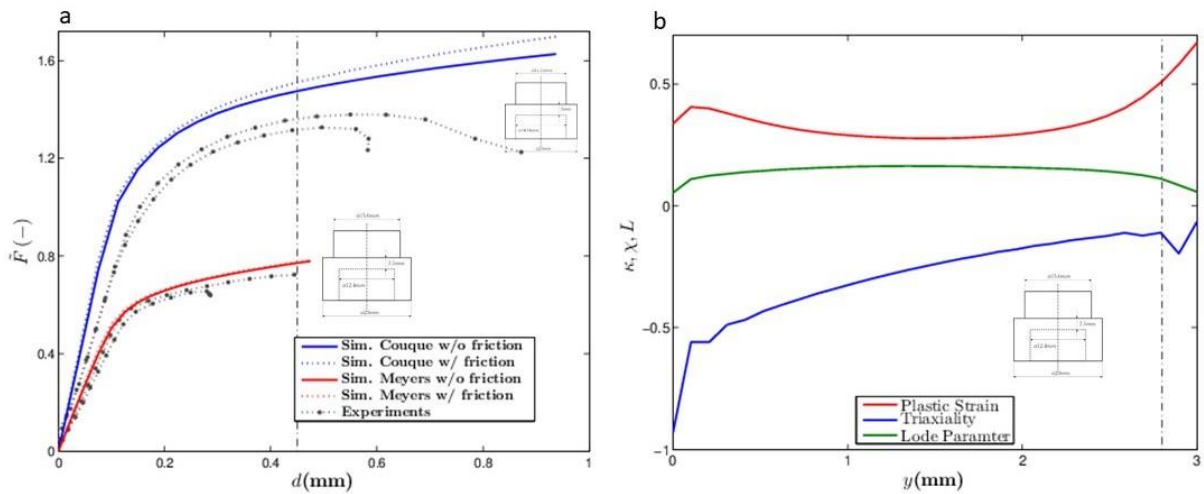
As visible in Figure 22b, as the mesh is refined in the sheared area, an oscillatory behaviour appears, similar to the one observed in the simulation of dynamic tension on the smooth specimen, in Figure 20d. The reason might be the mesh heterogeneity between the sheared area (fine mesh) and the surrounding part (coarser mesh) producing undesired stress discontinuities. Though the oscillatory response is visible on the overall reaction force at the head of the specimen, the plastic strain  $\kappa$ , the stress triaxiality ratio  $\chi$  and the Lode parameter  $L$  do not present such behaviour, as observed on the profiles computed along the vertical line joining the two notch roots in Figure 22c. The predicted values are only slightly dependent on the element size, except maybe for the plastic strain, for which a too rough mesh produces larger errors. It is assumed that an element size of  $100 \mu\text{m}$  is enough for a converged solution.

Moreover, although plastic strain eventually develops at the notch roots, the largest strain remains localized near the centre of the sheared area. Therefore, assuming a good correlation with the experimental strain fields, it is expected that damage and fracture would initiate near the centre.

### 3.4. Compression-shear loading

592 Figure 23a compares the measured and simulated normalized load-displacement curves for the  
 593 hat-shaped specimens loaded along ND. For the latter, two extreme boundary conditions were  
 594 considered: either frictionless contact between the sample and the holding device, which seems  
 595 reasonable, since some lubricant was used, or fully constrained sliding displacement on the  
 596 outer surface. For Meyer's samples ( $STR \approx -0.2$ ), the choice of one or the other boundary  
 597 condition has a negligible effect, and a reasonable agreement between simulations and tests can  
 598 be observed. Figure 23b shows the profiles of the plastic strain  $\kappa$ , stress triaxiality  $\chi$  and the  
 599 Lode parameter  $L$  along the blue line plotted in Figure 19c. For Couque's samples ( $\chi \approx -0.5$ ),  
 600 the effect of free or constrained sliding is slightly more pronounced, but the measured response  
 601 is not accurately captured, whatever the boundary condition. The proposed constitutive model  
 602 needs to be improved to reproduce the response of the material under such negative stress  
 603 triaxiality.

604



605

606 Figure 23: a) Measured and simulated normalized load-displacement curves for hat-shaped  
 607 compression-shear specimens,  $v = 8 \times 10^{-3} \text{ mms}^{-1}$ , loaded along ND, b) profiles of plastic strain  
 608  $\kappa$ , stress triaxiality  $\chi$  and Lode parameter  $L$  profiles along the blue line plotted in Figure 19c,  
 609 at the instant when fracture is expected to initiate, indicated by a vertical line in Figure 23a

610

### 611 3.5 Fracture loci

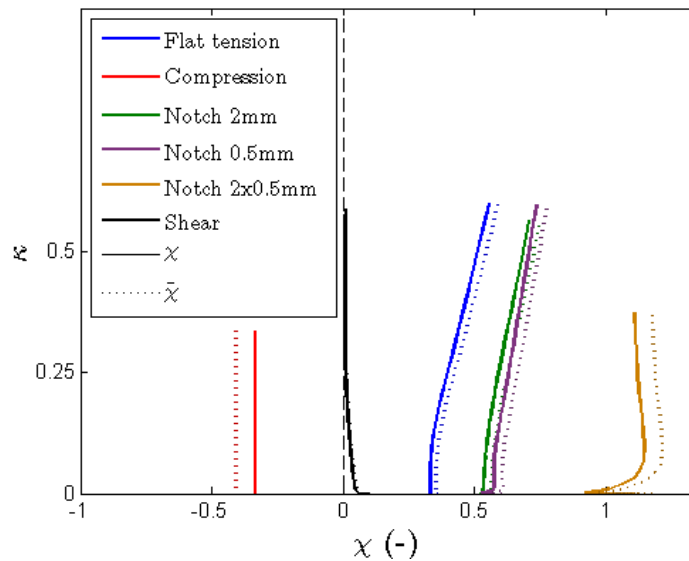
612

613 The local loading paths leading to fracture may be described through the evolutions of the stress  
 614 triaxiality ratio  $\chi$  and Lode parameter  $L$  that the different specimens have undergone at their  
 615 most critical point along plastic deformation. It provides an insight on how damage and  
 616 subsequent fracture occur as a function of the stress state.

617

618 Figure 24 shows the different loading paths computed with finite elements of the S/R type with  
 619 a size of approximately  $100 \mu\text{m}$ . The element whose loading path has been plotted is the  
 620 innermost element in the cross-section of the considered specimens, i.e. the furthest point from  
 621 the notches or edges. This choice is based on the assumption that the highest  $\chi$  is found away  
 622 from any free surface. However, for the notched geometry of radius  $R = 0.5\text{mm}$ , the loading  
 623 path were computed at an element close to the notches (as justified by the microscopic

624 observations and corroborated below).



625  
 626 Figure 24: Loading paths at the critical point for the different geometries in terms of plastic strain  
 627  $\kappa$  as a function of the stress triaxiality ratio  $\chi$ .  $\dot{\epsilon} \approx 10^{-3} s^{-1}$ , loading along TD. The two definitions  
 628 for the stress stress triaxiality ratios involve either Von Mises stress or the equivalent stress issued  
 629 from the constitutive model described in appendix

630  
 631 Depending on the definition used for the stress triaxiality ratio (using Von Mises stress, or an  
 632 equivalent stress related to the anisotropic constitutive model, equation 9 in the appendix) as  
 633 explained in the introduction, the loading paths may appreciably change. For instance,  $\tilde{\chi}$  in  
 634 compression is slightly more negative than  $\chi$  ( $\tilde{\chi} < \chi \approx -1/3$ ). Conversely, the tension  
 635 specimens present higher values of  $\tilde{\chi}$ . However, no large discrepancies are seen for the shear  
 636 conditions. Based on a comparison with values available in literature, the first definition of  $\chi$   
 637 based on Von Mises stress is retained in the sequel.

638  
 639 According to Figure 24, the stress triaxialities  $\chi$  are comparatively more positive in the notched  
 640 specimens than for the smooth samples, as expected. Both simple notches of  $R = 2mm$  ( $\chi \approx$   
 641  $1/2$ ) and  $R = 0.5mm$  ( $\chi \approx 2/3$ ) yield relatively close results. On the other hand, the doubly  
 642 notched specimen reaches a much higher positive triaxiality ratio ( $\chi > 1$ ). Surprisingly enough,  
 643 after some initial plastic deformation, the stress triaxiality reaches a steady value close to 1.2.  
 644 Finally, the tension-induced shear specimen ( $\chi \approx 0$ ) deforms under nearly pure shear stress  
 645 ( $\chi = 0$ ).

646  
 647 Figure 25 displays an estimation of the plastic strain value at which the different specimens fail  
 648 (represented by a dot) superimposed on the loading path. In order to find these values, the global  
 649 strain at fracture measured with the extensometer is compared with the one obtained from the  
 650 virtual extensometer in the simulation. The global strain at which the specimen is supposed to  
 651 fail is used to extract the local plastic strain in the finite element of interest. Table 4 gathers the  
 652 final values of  $\kappa$ ,  $\chi$  and  $L$  for all the successfully simulated types of tests, considering the latest  
 653 fracture among duplicated tests.

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**Table 4:** Local plastic strain at fracture, stress triaxiality and Lode parameter at fracture computed at the critical point. The values for Couque’s specimen are approximate (see above)

	Specimen type	Fracture strain, $\kappa$	Stress triaxiality, $\chi$	Lode parameter, $L$
RD	Smooth compression $10^{-3} \text{ s}^{-1}$	0.25	-0.33	1
	Shear	0.36	0.01	0
	Smooth tension $10^{-3} \text{ s}^{-1}$	0.2	0.38	-0.94
	Smooth tension $300 \text{ s}^{-1}$	0.14	0.33	-0.99
	Notched 2 mm	0.25	0.6	-0.15
	Notched 0.5 mm	0.2	0.62	-0.26
	Doubly notched	0.1	1.17	-0.07
TD	Smooth compression $10^{-3} \text{ s}^{-1}$	0.16	-0.33	1
	Shear	0.39	0.01	0.01
	Smooth tension $10^{-3} \text{ s}^{-1}$	0.25	0.4	-0.9
	Smooth tension $300 \text{ s}^{-1}$	0.13	0.33	-0.99
	Notched 2 mm	0.26	0.6	-0.16
	Notched 0.5 mm	0.2	0.61	0.27
	Doubly notched	0.1	1.15	-0.07
DD	Smooth compression $10^{-3} \text{ s}^{-1}$	0.28	-0.33	1
	Shear	0.44	0.03	0
	Smooth tension $10^{-3} \text{ s}^{-1}$	0.44	0.49	-0.8
	Smooth tension $300 \text{ s}^{-1}$	0.18	0.34	-0.98
	Notched 2 mm	0.27	0.6	-0.1
	Notched 0.5 mm	0.27	0.67	-0.22
	Doubly notched	0.12	1.15	0
ND	Couque $10^{-3} \text{ s}^{-1}$ (approximations)	$\approx 0.4$	$\approx -0.75$	$\approx 0.1$
	Meyer $10^{-3} \text{ s}^{-1}$	0.67	-0.43	0.1
	Smooth compression $10^{-3} \text{ s}^{-1}$	0.38	-0.33	1
	Smooth tension $10^{-3} \text{ s}^{-1}$	0.15	0.36	-0.93

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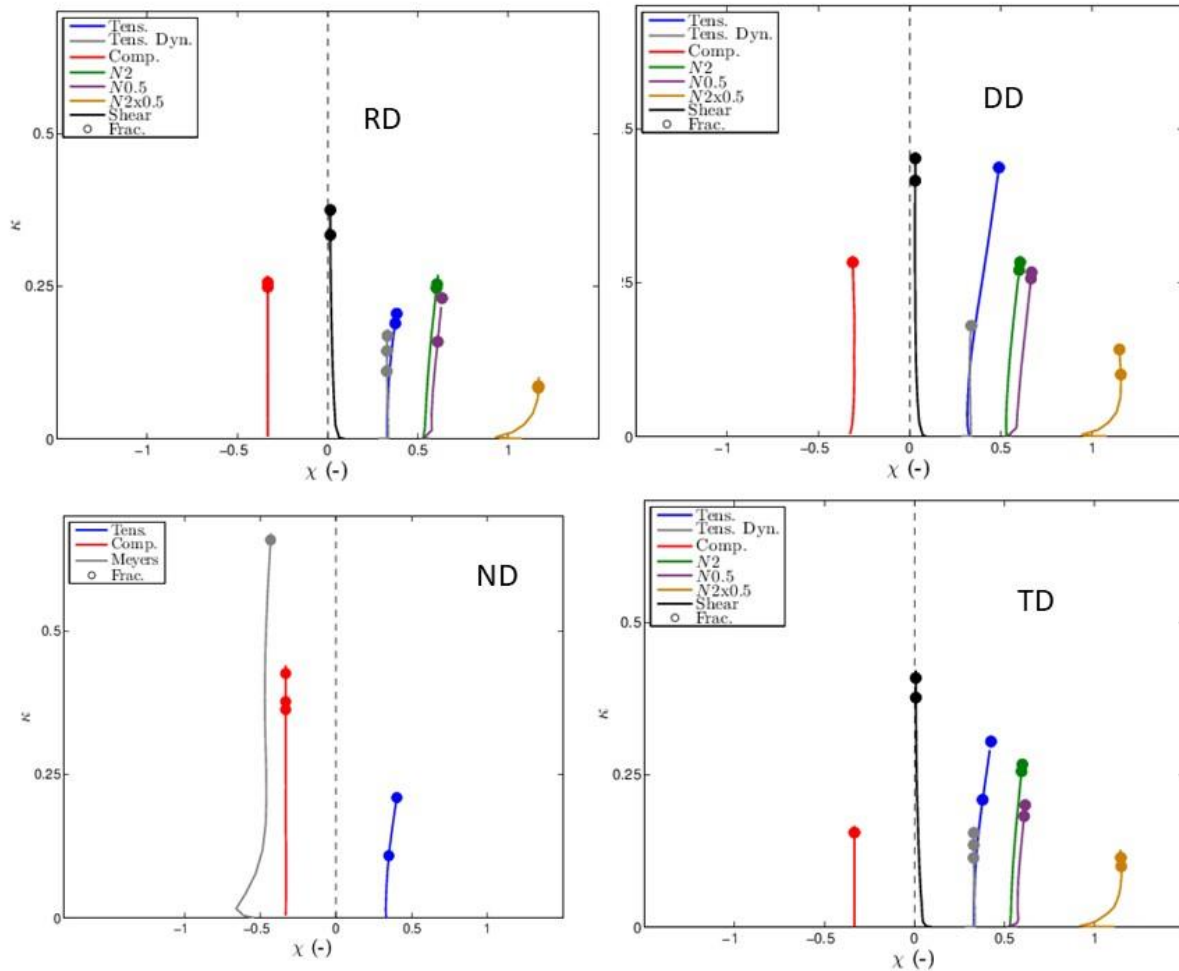


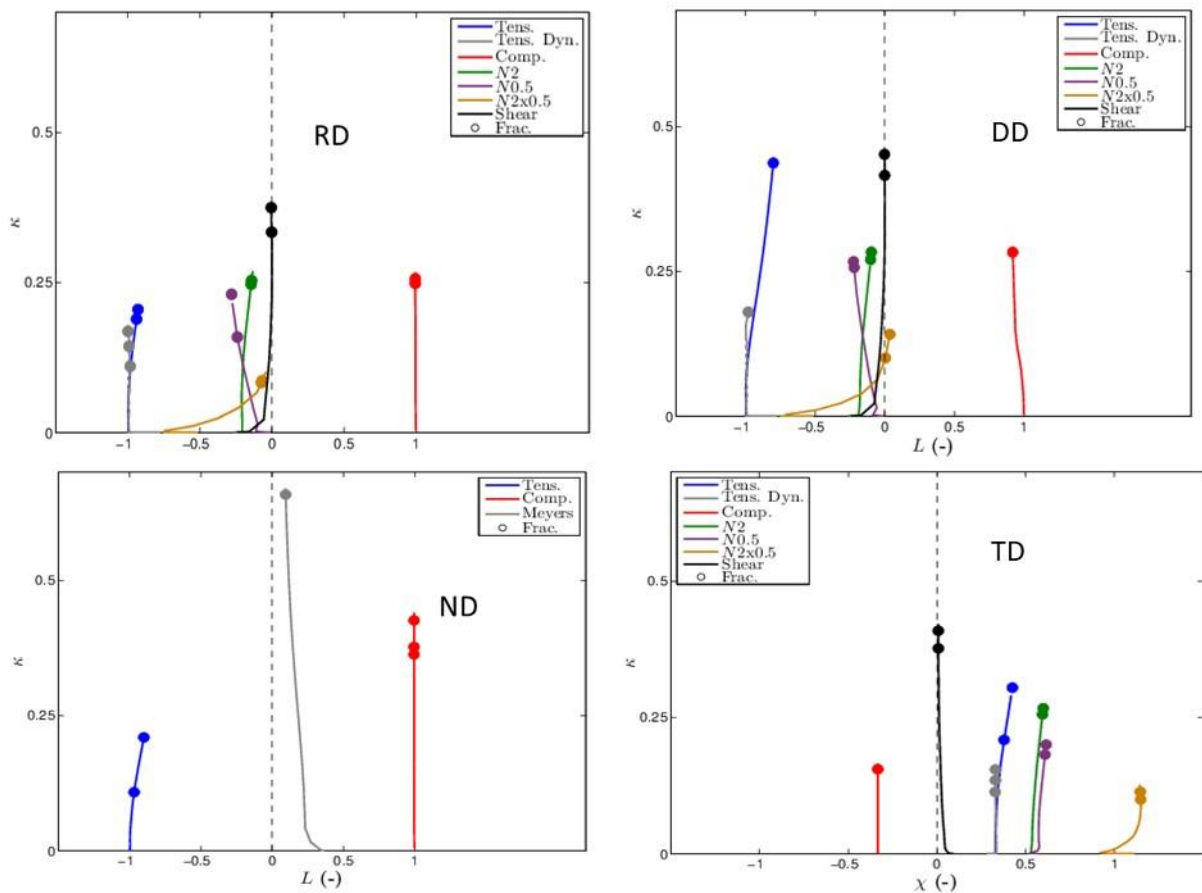
Figure 25: Plastic strain  $\kappa$  vs. stress triaxiality ratio  $\chi$  for all loading directions. The dots indicate fracture.

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Figure 25 suggests a smaller strain to fracture when the triaxiality increases in absolute values, with a maximum ductility for  $\chi$  close to zero, except along ND where the peak ductility seems to be shifted towards a negative triaxiality. This result is appreciably different from the fracture locus obtained for AA2024 by Bao and Wierzbicki(2004) . This result is however consistent with the evolution found by Papisidero et al. (2015) for the same aluminium alloy, and with the conclusion of a ductility higher in shear than in tension drawn by Gross et al. (2016) as well as by Wang et al. (2019) for Ti-6Al-4V. Regarding the strictly positive stress triaxiality range, the strain to fracture decreases as the stress triaxiality  $\chi$  becomes more positive. This expected result is in accordance with Rice and Tracey’s (1969) and Johnson-Cook’s (1985) fracture loci. The lower ductility found here for a small negative triaxiality, as compared to pure shear – at least for loadings along RD, TD and DD – seems actually in agreement with the results of Vilotic et al. for Ti-6Al-4V(2016) , since their Figure 6 shows a fracture strain for  $\chi$  close to -0.4, lower than for  $\chi = 0$ , and then a rise in ductility, as the triaxiality gets more negative. Additional tests in such conditions would be useful to confirm and explain this tendency. The evolution of the local plastic strain  $\kappa$  as a function of the local Lode parameter  $L$  at the same elements as in Figure 25 is plotted in Figure 26. No clear trend emerges from the latter,

679 except a peak of ductility for a Lode parameter close to 0.

680



681

682 Figure 26: Plastic strain  $\kappa$  vs. Lode parameter,  $L$  for all loading directions. The dots indicate  
683 fracture.

#### 684 4 Concluding remarks

685

686 A series of mechanical tests have been carried out in the stress triaxiality range  $[-1/2,1]$  and for  
687 strain rates ranging from  $10^{-3}$  to  $1500 \text{ s}^{-1}$  on a Ti-6Al-4V titanium alloy along various loading  
688 directions. Microscopic observations of broken specimens have been achieved in order to  
689 document the underlying damage mechanisms. While the growth and coalescence of cavities  
690 (nucleated mostly at  $\alpha/\beta$  interfaces) control ductile fracture at positive triaxialities, strain  
691 localization, and shear mode cracking dominate at zero and negative triaxialities.

692

693 Given the limited micro-damage and the rather brutal drop in stress at failure, a fracture criterion  
694 considering a damage indicator seems to be suitable for the Ti-6Al-4V titanium alloy under  
695 consideration. Yet, another batch (with e.g. a different heat treatment) may potentially fail  
696 differently and possibly require the use of a coupled damage-plasticity model, see Longère and  
697 Dragon (2015).

698

699 In the present case, the plastic strain at failure is weakly dependent on the strain rate in the  
700 considered range, and may be expressed in a first approximation via an exponentially



701 decreasing function of the absolute value of the stress triaxiality ratio. This function depends  
 702 significantly on the loading direction, with nearly a factor of two between the most and least  
 703 ductile directions. These directions change between tension and compression, probably because  
 704 of changes in the contribution of mechanical twinning. This tension-compression asymmetry in  
 705 terms of anisotropy in ductility makes the formulation of an anisotropic fracture criterion  
 706 difficult.

707  
 708 By contrast, no clear dependence on the Lode parameter or strain rate was found.  
 709 Accordingly, the fracture criterion proposed by Bao and Wierzbicki (2004) and Wierzbicki et  
 710 al. (2005) must not be considered as universal, and there actually exists no universal cut-off  
 711 value for negative stress triaxiality, viz.  $-1/3$ . Indeed, this cut-off value was derived from a series  
 712 of tests carried out on a very specific class of materials (AA2024-T351 aluminum alloys) and  
 713 is not physically supported for others. On the contrary, there are many experimental evidences  
 714 in literature of failure occurring under STR lower than  $-1/3$ , (see e.g. Longère and Dragon  
 715 (2015) or Vilotic et al. (2016) ), so that shear decohesion is not bounded in the negative stress  
 716 triaxiality range by a value of  $-1/3$ . In addition, an overall nil (and even negative) stress triaxiality  
 717 at the macroscale does not necessarily imply that the stress triaxiality is nil or negative  
 718 everywhere at the micro-scale, as shown in Longère and Dragon (2015), which explains the co-  
 719 existence of smooth, sheared areas and dimple clusters in specimens after shear-compression  
 720 loading until fracture.

721  
 722 The readers interested in attenuating the mesh dependence of the numerical results in presence  
 723 of damage-induced softening can refer to Abu Al-Rub and Voyiadjis (2006), Voyiadjis and Abu  
 724 Al-Rub (2006), and Allix (2012).

725  
 726  
 727 **Appendix: Anisotropic viscoplastic Constitutive model**

728  
 729  
 730 The viscoplastic, anisotropic constitutive model specifically formulated and identified for this  
 731 batch of Ti-6Al-4V titanium alloy is detailed in Ruiz de Sotro et al. (2020-a) . It accounts for  
 732 plastic orthotropy, isotropic and kinematic hardening, tension-compression strength differential  
 733 and viscosity, and is only briefly introduced in the following.

734 In the constitutive model, the set of state variables  $\wp$  and conjugate state variables  $\wp^c$  are:

735  
 736 
$$\wp = (T, \underline{\underline{\varepsilon}}^e, \kappa, \underline{\underline{a}}) \Leftrightarrow \wp^c = (s, \underline{\underline{\sigma}}, r, \underline{\underline{X}}) \quad (5)$$

737  
 738 where  $(T, \underline{\underline{\varepsilon}}^e, \kappa, \underline{\underline{a}})$  represents temperature, elastic strain  $2^{\text{nd}}$  order tensor, isotropic hardening  
 739 variable (and cumulated plastic strain) and kinematic hardening variable  $2^{\text{nd}}$  order tensor, and  
 740  $(s, \underline{\underline{\sigma}}, r, \underline{\underline{X}})$  entropy, Cauchy stress  $2^{\text{nd}}$  order tensor, isotropic hardening force and kinematic  
 741 hardening force  $2^{\text{nd}}$  order tensor.



742 The rate form of the constitutive state laws in the corotational frame reads:

743

$$744 \quad \underline{\underline{\dot{\sigma}}} = \underline{\underline{C}} : \underline{\underline{\dot{\varepsilon}}}^e - \alpha K \dot{T} \underline{\underline{\delta}} ; \underline{\underline{\dot{X}}} = \frac{2}{3} C \dot{a} ; \dot{r} = h''(\kappa) g(T) \dot{\kappa} + h'(\kappa) g'(T) \dot{T} \quad (6)$$

745

746 where  $(\alpha, K, C)$  represent the thermal dilatation coefficient, bulk modulus and kinematic

747 hardening related constant. The stiffness 4<sup>th</sup> order tensor  $\underline{\underline{C}}$ , stored energy of cold work  $h(\kappa)$

748 , and thermal softening function  $g(T)$  are expressed as:

749

$$750 \quad \begin{cases} \underline{\underline{C}} = \lambda \underline{\underline{\delta}} \otimes \underline{\underline{\delta}} + 2\mu \underline{\underline{I}} \\ h(\kappa) = \frac{Q_0}{n+1} (\varepsilon_0 + \kappa)^{n+1} - \frac{C}{D} [D\kappa - \exp(-D\kappa)] \\ g(T) = 1 - \left\langle \frac{T - T_{ref}}{T_m - T_{ref}} \right\rangle^{m_r} \end{cases} \quad (7)$$

751

752 where  $(\lambda, \mu)$  are Lamé elasticity constants and  $(Q_0, n, \varepsilon_0, D)$  isotropic hardening related

753 constants.  
754 The yield function  $f(\underline{\underline{\hat{\Sigma}}}, r; k)$  and plastic potential  $F(\underline{\underline{\hat{\Sigma}}}, r; k)$  accounting for rate dependent,

755 non-associated plasticity read:

$$756 \quad \begin{cases} f(\underline{\underline{\hat{\Sigma}}}, r; k) = \sigma_{eq}(\underline{\underline{\hat{\Sigma}}}_p; k) - [R_0 g(T) + r(\kappa, T)] = \sigma_v(\dot{\kappa}) \geq 0 \\ F(\underline{\underline{\hat{\Sigma}}}, r; k) = f(\underline{\underline{\hat{\Sigma}}}, r; k) + \frac{3D}{4C} \underline{\underline{X}} : \underline{\underline{X}} \end{cases} \quad (8)$$

758

759 where  $R_0$  is a constant ( $R_0 g(T_0)$  represents the initial size of the elasticity domain at

760 temperature  $T_0$ ) and  $\sigma_v(\dot{\kappa})$  the viscous stress. The transformed equivalent stress  $\sigma_{eq}(\underline{\underline{\hat{\Sigma}}}_p; k)$  is

761 expressed as:

762

$$763 \quad \begin{cases} \sigma_{eq}^a(\underline{\underline{\hat{\Sigma}}}_p; k) = \frac{1}{m_0^a} \sum_{p=1}^3 \left( |\hat{\Sigma}_p| - k \hat{\Sigma}_p \right)^a ; \hat{\Sigma}_p = \text{eig}_p(\underline{\underline{\hat{\Sigma}}}) ; \underline{\underline{\hat{\Sigma}}} = \underline{\underline{A}} : (\underline{\underline{s}} - \underline{\underline{X}}) \\ m_0^a = \left[ \frac{2}{3} (1 - |k|) \right]^a + 2 \left[ \frac{1}{3} (1 + |k|) \right]^a \end{cases} \quad (9)$$

764

765 where the constants  $(k, a)$  allow for describing the tension-compression asymmetry and shape

766 of the elasticity domain.  $\text{eig}_p(\hat{\underline{\underline{\Sigma}}})$  is the p-th eigenvalue of  $\hat{\underline{\underline{\Sigma}}}$ . Voigt notation for  $\underline{\underline{A}}$  yields the  
 767 following 6x6 matrix representation

$$768 \quad [A] = \begin{bmatrix} A_{11} & A_{12} & A_{13} & & & \\ A_{12} & A_{22} & A_{23} & & & \\ A_{13} & A_{23} & A_{33} & & & \\ & & & A_{44} & & \\ & & & & A_{55} & \\ & & & & & A_{66} \end{bmatrix} \quad (10)$$

770 where the  $A_{ij}$  terms are constants.

771 According to the normality rule, the evolution laws reads in the corotational frame:

$$772 \quad \underline{\underline{\dot{\varepsilon}}}^p = \dot{\kappa} \underline{\underline{n}} ; \underline{\underline{\dot{a}}} = \underline{\underline{\dot{\varepsilon}}}^p - D \dot{\kappa} \underline{\underline{a}} ; \dot{T} = \frac{\beta \sigma_{vm}}{\rho c} \dot{\kappa} \quad (11)$$

773 where  $(\underline{\underline{\dot{\varepsilon}}}^p, \underline{\underline{n}})$  are the plastic strain rate and yield direction 2<sup>nd</sup> order tensors, and  $(\beta, \rho, c)$  the  
 774 constitutive model consistent inelastic heat fraction (proportion of plastic work rate converted  
 775 into heat), mass density and specific heat. One has also:

$$776 \quad \left\{ \begin{array}{l} \dot{\kappa} = \left\langle \frac{f}{Y_v} \right\rangle^{n_v} ; \sigma_v(\dot{\kappa}) = Y_v \dot{\kappa}^{1/n_v} \\ n = \frac{1}{m_0^a} \sum_{p=1}^3 \left( \frac{|\hat{\underline{\underline{\Sigma}}}_p| - k \hat{\underline{\underline{\Sigma}}}_p}{\sigma_{eq}} \right)^{a-1} \left( \text{sgn}(\hat{\underline{\underline{\Sigma}}}_p) - k \right) \underline{\underline{J}} : \underline{\underline{A}} (\underline{\underline{v}}_p \otimes \underline{\underline{v}}_p) ; \underline{\underline{J}} = \underline{\underline{I}} - \frac{1}{3} \underline{\underline{\delta}} \otimes \underline{\underline{\delta}} \\ \beta = \frac{\sigma_{eq} - r + \frac{2}{3} CD a : a}{\sigma_{vm}} ; \sigma_{vm} = \sqrt{\frac{3}{2} s : s} \end{array} \right. \quad (12)$$

781 For confidentiality reason, the values of the material constants are not provided.

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