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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

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 χ , expected to control micro-
 22 void collapse/growth, and the shear stress ratio η , expected to control micro-void flattening and
 23 subsequent shear decohesion, or the Lode angle θ 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 σ_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 ε_F^p or stress at
 37 fracture σ_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 ε_F^p on χ in the $[0;1.2]$ χ -
 40 range (as well as on strain rate and temperature). Bao and Wierzbicki (2004) evidenced a non-
 41 monotonic dependence of ε_F^p on χ in the $[-1/3;1]$ χ -range for AA2024-T351 aluminum alloy,
 42 and accordingly Bao and Wierzbicki (2005) proposed a 2-regime based dependence of ε_F^p on
 43 χ . In the same time, Wierzbicki and Bao (2005) reported a criterion accounting explicitly for
 44 χ and L in the expression of ε_F^p . For the same aluminum alloy for which they obtain different
 45 fracture conditions in the $[0;1]$ χ -range, i.e. a monotonic decrease of ductility, Papasidero et
 46 al. (2015) identified a θ - and loading direction-dependent σ_F (and further ε_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 $[0;2]$ χ -range, Hooputra et al. (2004)
 49 developed two competing ε_F^p related fracture criteria, involving χ and η , 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 χ 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 χ 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 σ_F leading to a χ - and L -dependence of
77 ε_F^p . The experimental scatter in ε_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⁻¹. They reported a
83 nearly constant ductility up to 400°C, and a substantial rise beyond. The anisotropy of ε_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 χ 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 2.5x10³ and 10⁵ s⁻¹ at room temperature or at
90 100°C. They reported that ε_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 χ 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 ε_F^p as χ 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 ε_F^p ,
102 assumed to decrease monotonically with χ 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 χ , 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 ε_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 χ 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 (χ 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

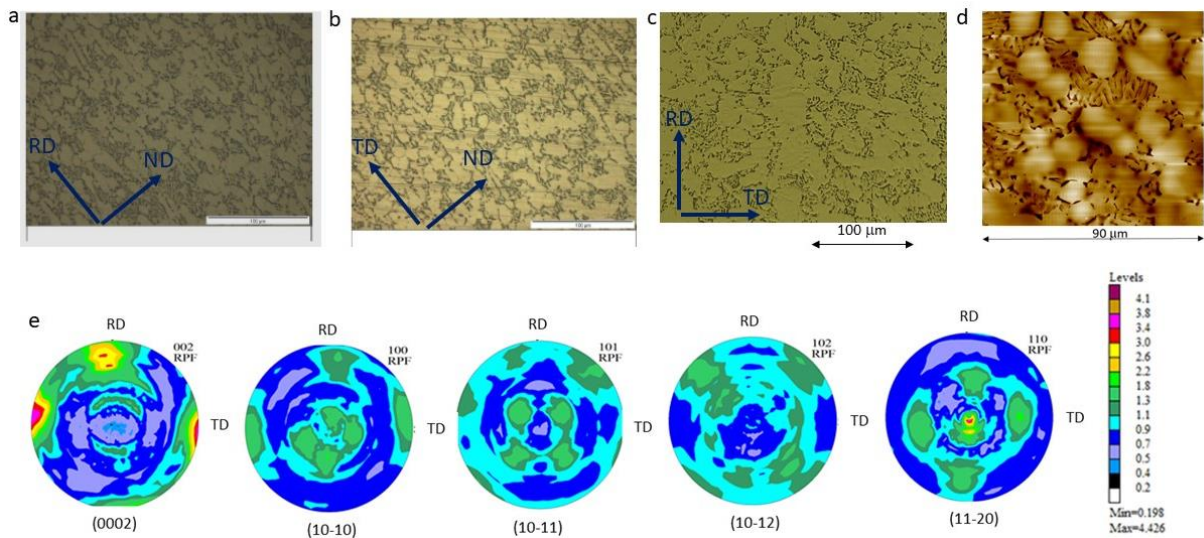
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136 2.1 Material under consideration and geometry of the specimens

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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 α phase nodules ranges from a few microns up to 30 μm . Secondary α laths, 5 to 20 μm
 141 long and a few μm wide are visible within the partially transformed β phase (dissolved by the
 142 chemical attack) on the atomic force microscopy (AFM) image (Figure 1d). The spatial
 143 distribution of the α and β phases does not seem completely random, since in many places, rows
 144 of α 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 β 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 μm wide "macro-zones" with a different orientation,
 149 inherited from the orientation of the former β 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 α 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 χ
 162 and L , viz. $\chi \in [-1,1.2]$ and $L \in [-1,1]$. As a reminder, Table 1 reports the values of χ , η
 163 and L for a few loading cases.
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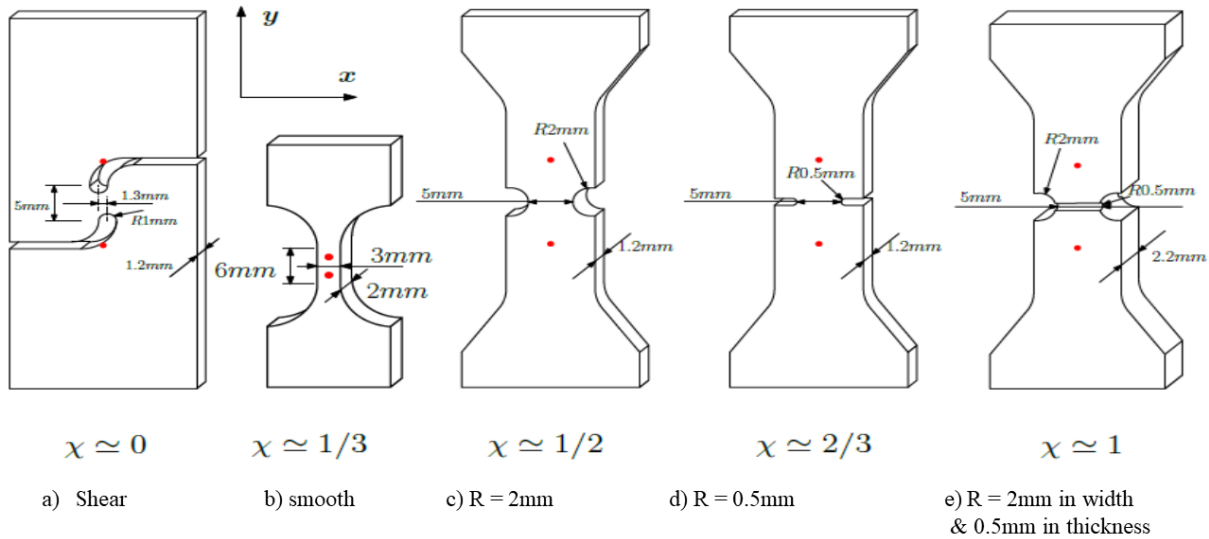
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Table 1: A few loading types and their corresponding values of stress triaxiality ratio χ , shear stress ratio η , and Lode parameter, L .

	σ_1	σ_2	σ_3	p	τ_{\max}	$q = \sigma_M$	χ	η	χ/η	L
Equi-biaxial compression	0	$-\sigma$	$-\sigma$	$2\sigma/3$	$\sigma/2$	σ	$-2/3$	$1/2$	$-4/3$	-1
Compression	0	0	$-\sigma$	$\sigma/3$	$\sigma/2$	σ	$-1/3$	$1/2$	$-2/3$	1
Shear	τ	0	$-\tau$	0	τ	$\sqrt{3}\tau$	0	$1/\sqrt{3}$	0	0
Tension	σ	0	0	$-\sigma/3$	$\sigma/2$	σ	$1/3$	$1/2$	$2/3$	-1
Equi-biaxial tension	σ	σ	0	$-2\sigma/3$	$\sigma/2$	σ	$2/3$	$1/2$	$4/3$	1

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Figure 2 shows the geometries of the specimens leading to nil and positive χ , with the tension-induced shear specimen inspired from the work of Roth and Mohr(2018), and Figure 3 those leading to negative χ 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 χ 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.

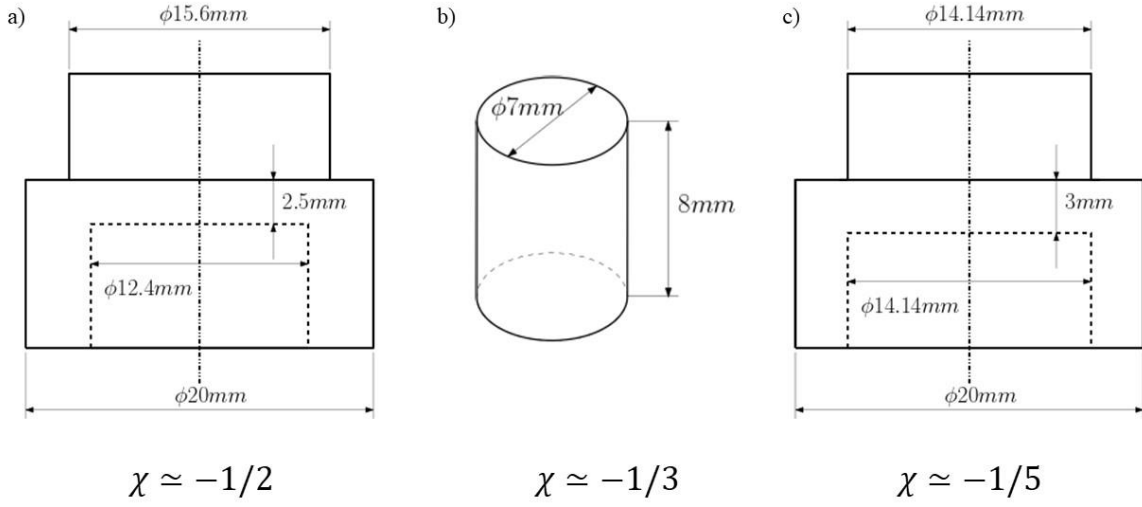


Figure 3: Axisymmetric specimen geometries used to cover the stress triaxiality range $\chi \in [-1/2, 0]$. Hat-shaped specimens have been designed following Couque (2003) and Meyers et al. (2001)

Tension, compression and shear tests were carried out at room temperature and at low strain rate (typically ranging between $|\dot{\epsilon}| \approx 10^{-4} s^{-1}$ and $|\dot{\epsilon}| \approx 10^{-1} s^{-1}$) using standard tension-compression testing machines, and at high strain rate (typically of the order of $|\dot{\epsilon}| \approx 1.5 \times 10^3 s^{-1}$) using split Hopkinson pressure bar (SHPB) set-ups. In order to assess repeatability, two to four test repetitions were carried out for most tests. In the case of compression tests, a clip-on extensometer was mounted on the rigid plates compressing the sample, and grease was applied as lubricant on the top and bottom sides in order to minimize the barreling effect. For dynamic tension tests, the load-inversion device developed by Dunand et al. (2013) and modified by Roth et al. (2015) was used. In these tests, the strain field in the sample was obtained by digital image correlation (DIC) using Vic-2D software, based on images captured by a Phantom v7.3 high speed camera. A frame rate of 10^5 Hz with a resolution of 304×64 pixel² was employed to monitor a 10×3 mm² area covered with speckle painting. The global strain in the gage length was measured using a virtual extensometer based on the distance between two dots on the specimen, see Figures 2-3. Table 2 summarizes the test conditions.

Table 2: Loading conditions

Specimen type	Loading direction	Strain or displacement rate
Shear	RD, TD, DD	1.6×10^{-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	1.6×10^{-3} mm/s
Simply notched 0.5mm	RD, TD, DD	1.6×10^{-3} mm/s
Doubly notched 0.5mm	RD, TD, DD	1.6×10^{-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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209 2.2 Mechanical characterization

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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} 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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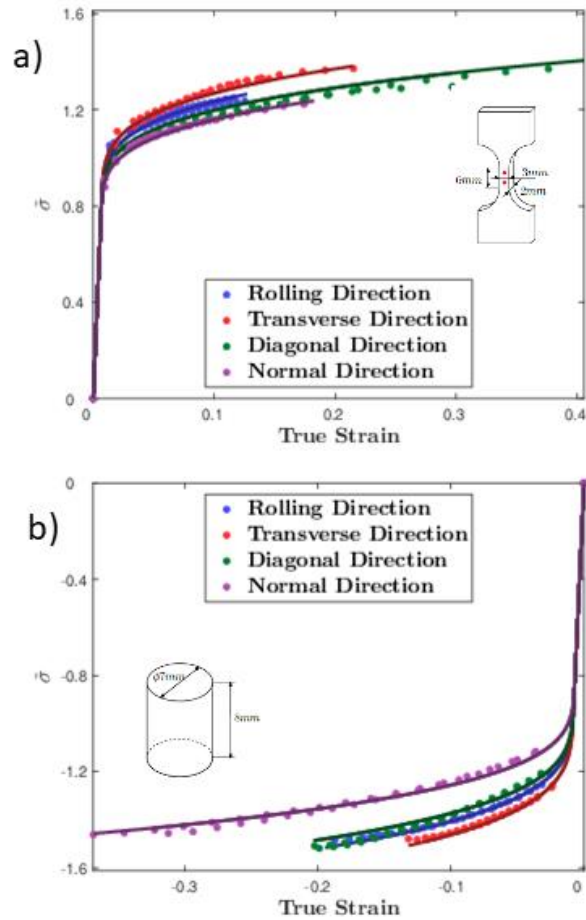
218 First, the influence of the loading direction and strain rate on the flow curves under uniaxial
 219 compression or tension is depicted in Figures 4 and 5. The end point of each curve corresponds
 220 to fracture. The curves obtained by simulation using the constitutive model outlined in
 221 Appendix 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 α 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.

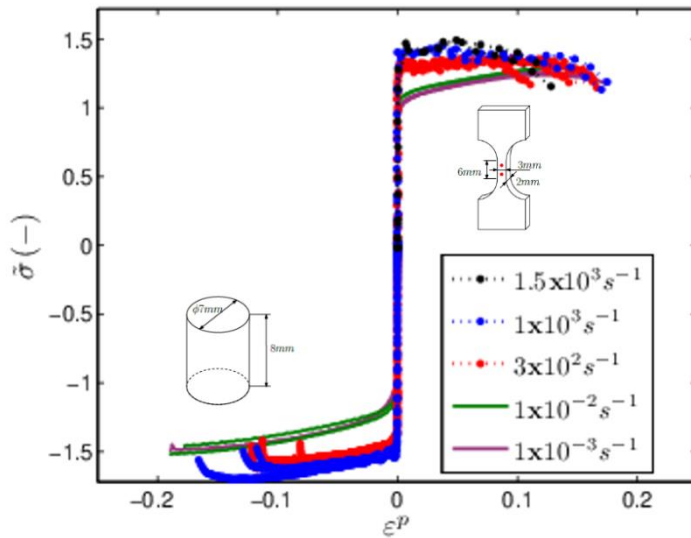
228 All these specificities, as well as a pronounced kinematic hardening (not documented here)
 229 were captured by the constitutive model detailed in Ruiz de Sotro et al. (2020-a) and more
 230 briefly recalled in Appendix.

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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 ϵ_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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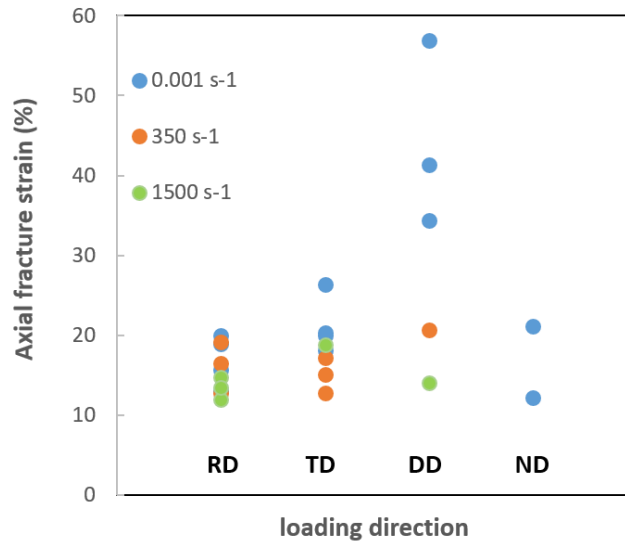


Figure 6: Variation of the axial tensile fracture strain with the loading direction and strain rate

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 characteristics 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).

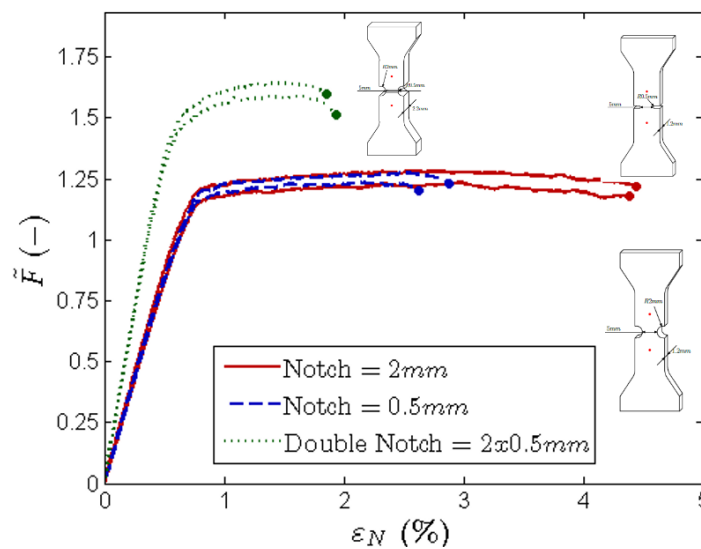
Contrary to what is observed on those smooth tensile specimens, for which the ductility along ND does not seem very different than along RD and TD, smooth cylindrical specimens compressed along ND failed at a much higher strain (38%) than along RD (25%), TD (16%) or even DD (28%). This is consistent 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 $10^{-3} s^{-1}$.

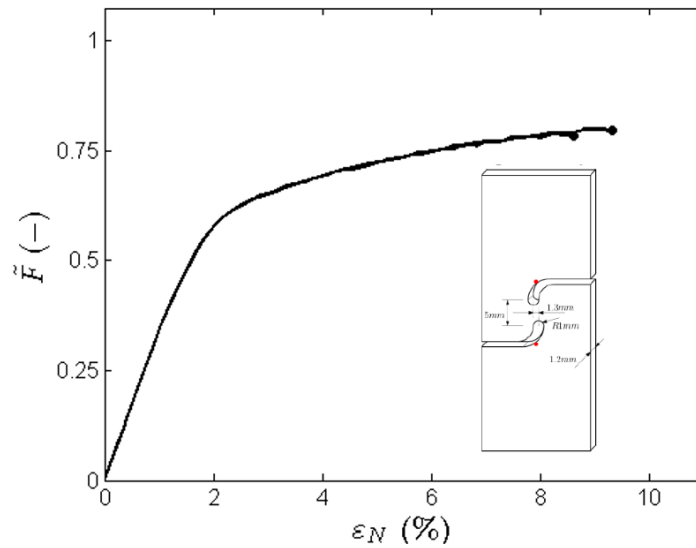
300 By order of importance of its effect on ductility, it seems that the loading direction with respect
 301 to the cold-rolled sheet (thickness as opposed to width) comes first, then the in-plane loading
 302 direction (RD, DD, TD), followed by the loading rate, for which no rule can be drawn.

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 χ . 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.



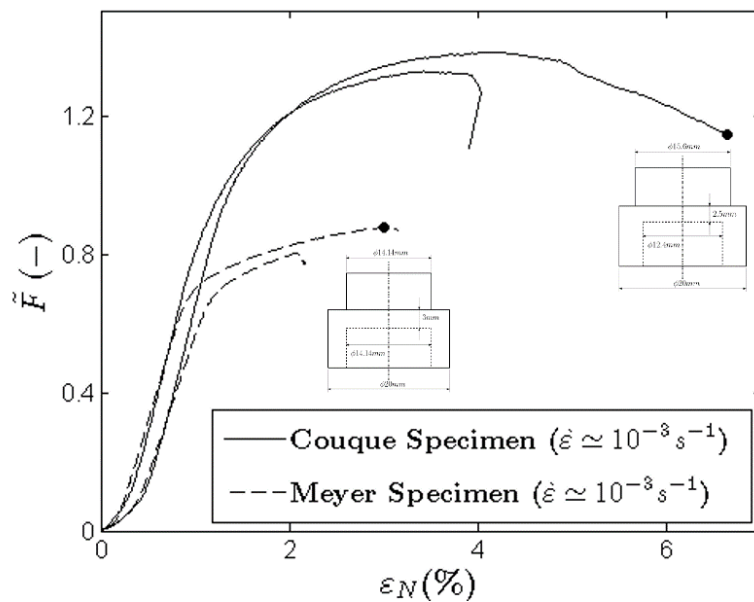
310
 311 Figure 7: Normalized load versus apparent strain for tensile tests along TD on notched
 312 specimens. Displacement rate: $1.6 \times 10^{-3} mms^{-1}$. Fracture is denoted by a circle
 313

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 χ .

Positive stress triaxiality χ

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 α and β phases, although a few very small cavities are found within primary α 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.

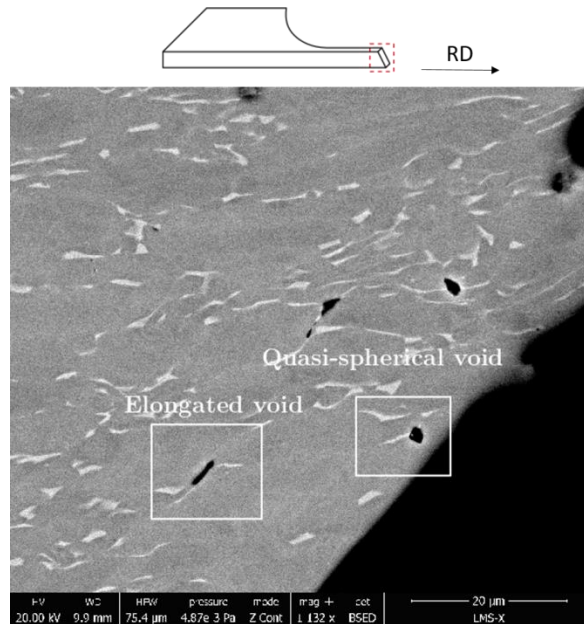
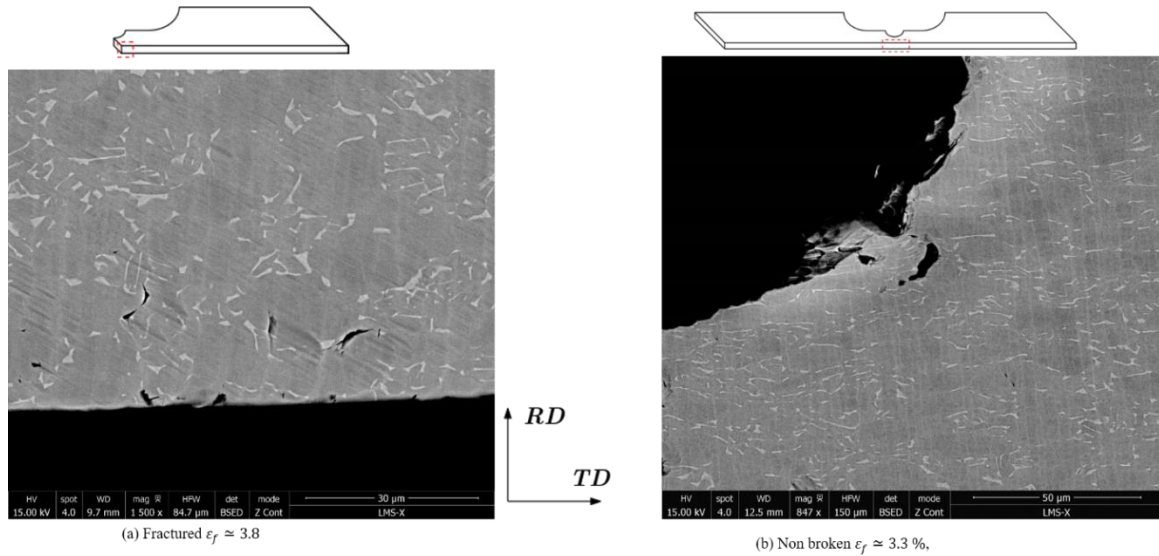


Figure 10: SEM observation of microvoids on a longitudinal section of a smooth tensile specimen fractured at 350 s^{-1} along RD

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° , 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 α nodules perpendicular to
 373 ND mentioned above.
 374

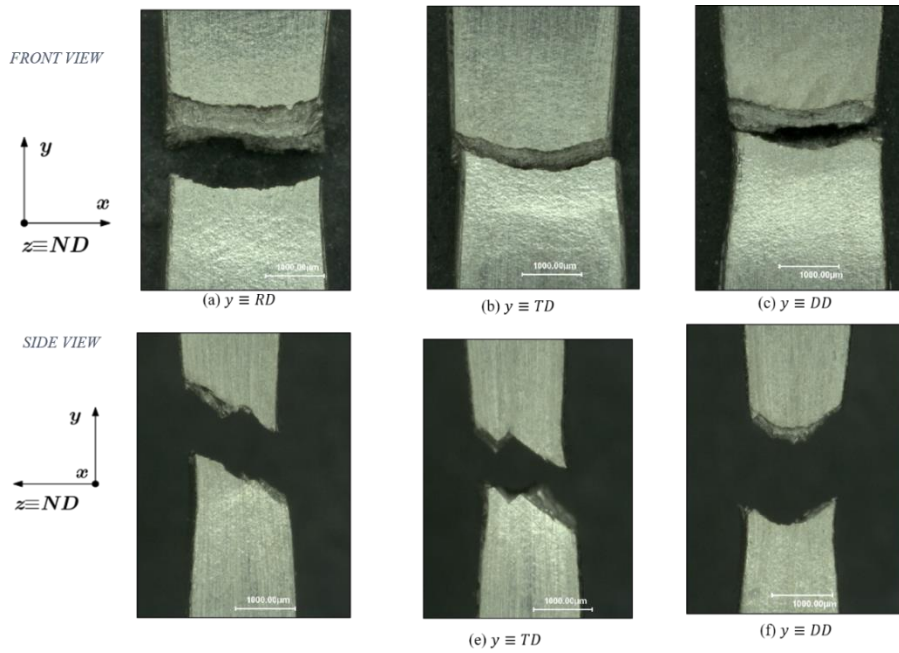


Figure 12: Broken smooth tensile specimens ($\chi \approx 1/3$). $\dot{\epsilon} = 10^{-3} \text{s}^{-1}$

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.

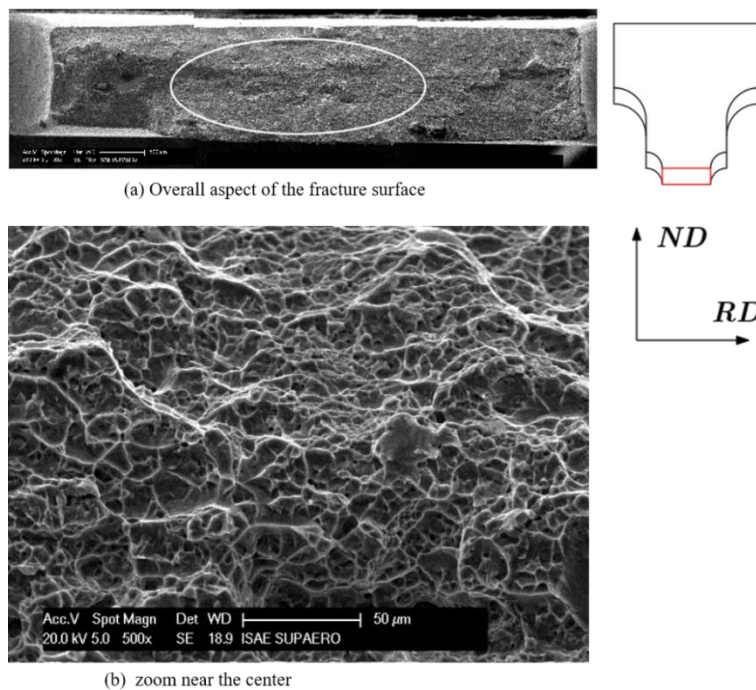


Figure 13: a) SEM micrograph of a notched TD specimen ($R=2 \text{ mm}$, $\chi \approx 1/2$). $\epsilon_f \approx 3.8 \%$, 10^{-3}

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 χ is
396 speculated to occur away from the center-line. This hypothesis will be confirmed by numerical
397 simulations presented in the following Section.

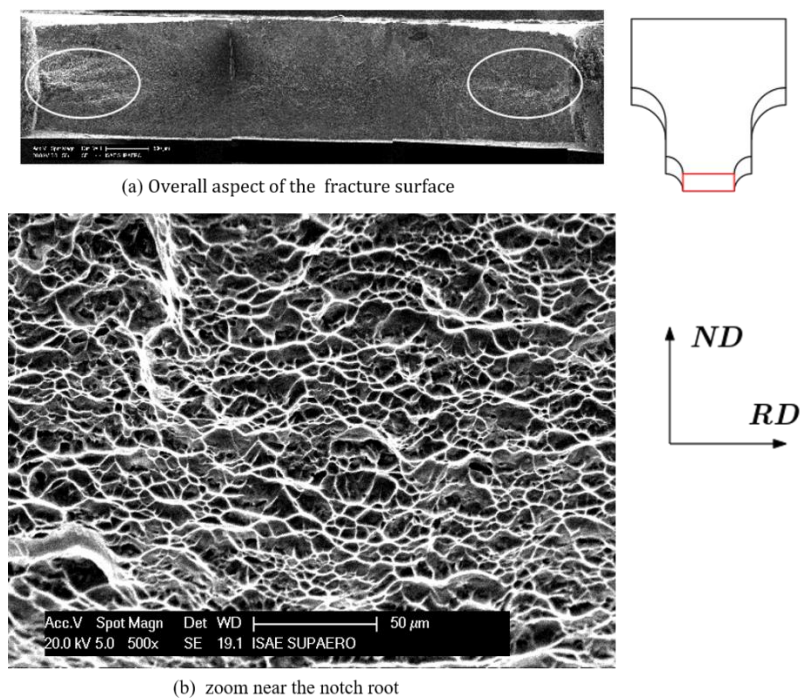
398

399 Nil stress triaxiality χ

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



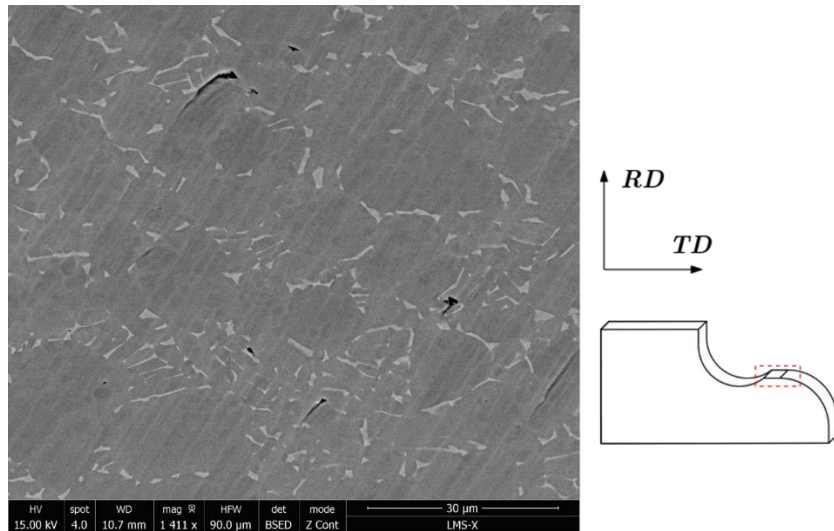
406

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

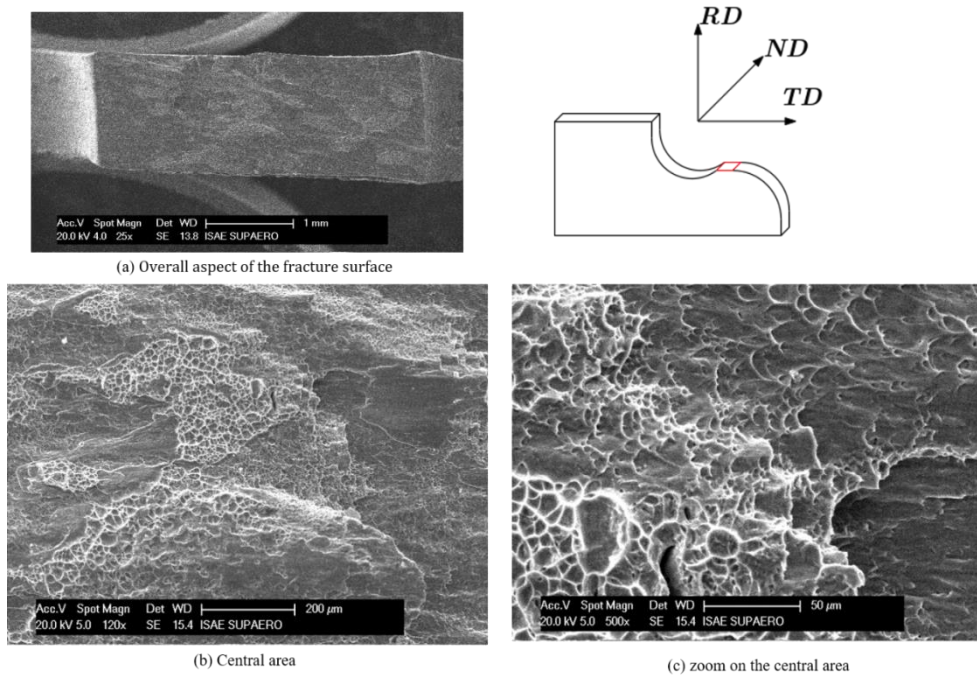
409

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}s^{-1} .
 421

422 Negative stress triaxiality χ

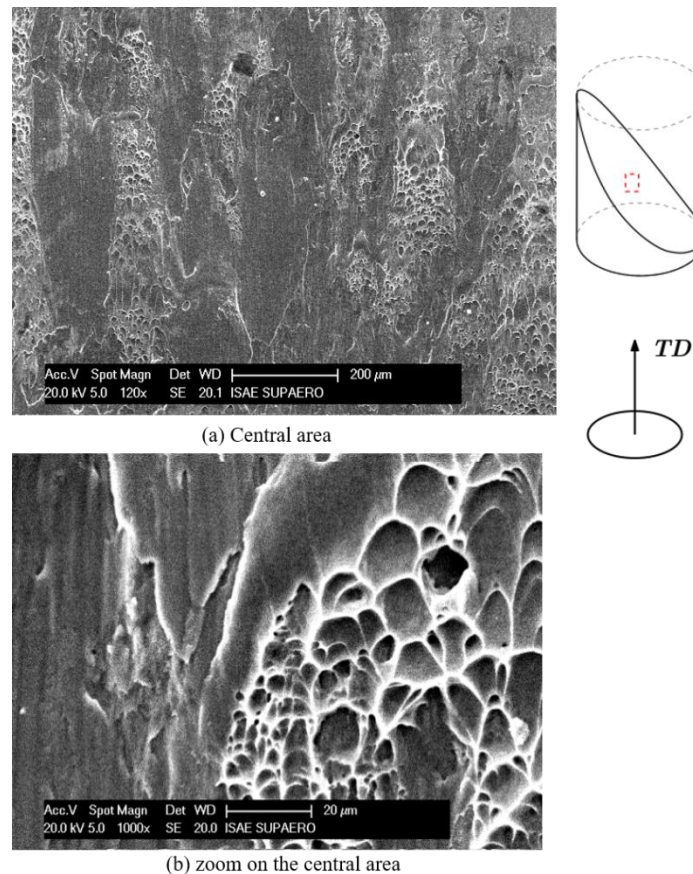
423

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

425 in a band inclined by 45° 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

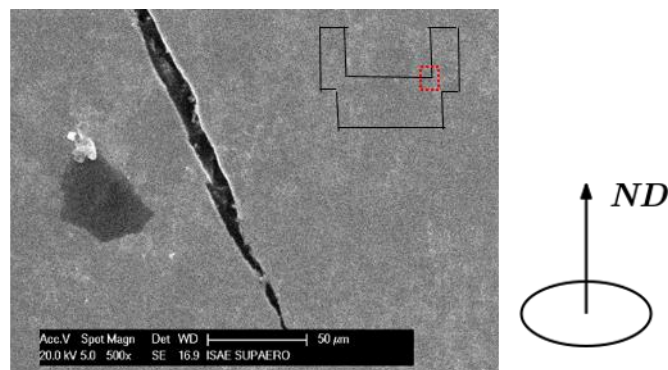
447

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 χ
 450 (compression and shear loadings), and (ii) a cavity nucleation/growth and coalescence
 451 mechanism, typical of positive χ . This competition between shear localization and micro-
 452 voiding must accordingly be accounted for in the formulation of a fracture criterion.

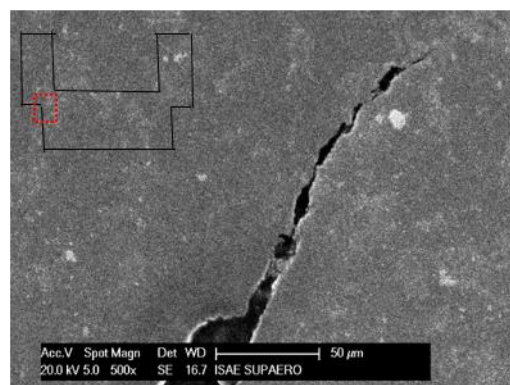
453 Damage, in the form of micro-cavities, nucleated mostly at α/β interfaces, was found near the
 454 fracture surfaces. The latter were often inclined by 45° into the thickness of smooth tension
 455 specimens. Depending on the loading conditions, shear-driven strain localization and damage,
 456 leading 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

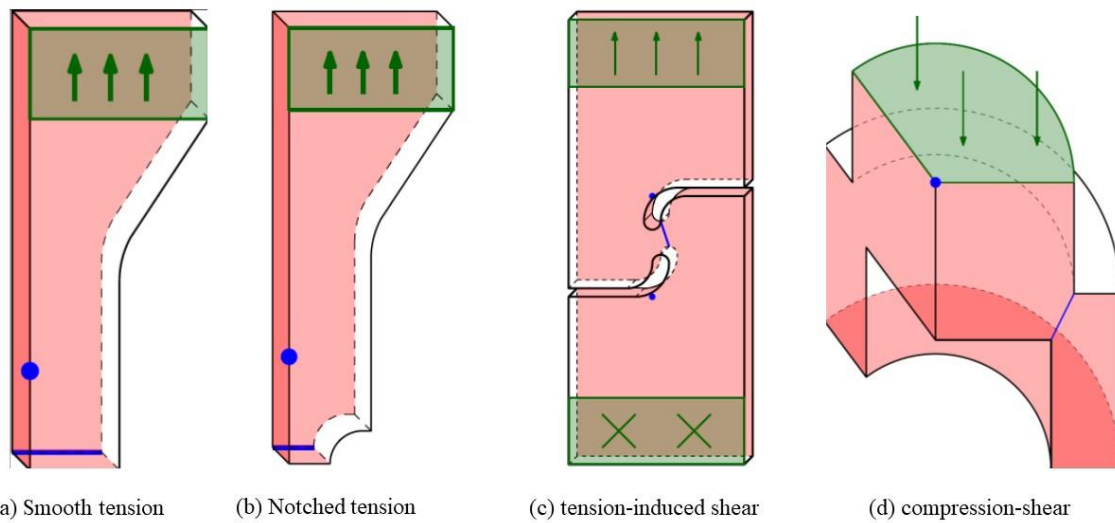
473 The constitutive model used to simulate the tests, detailed in Ruiz de Sotro et al. (2020-a) and
474 outlined in Appendix, does not account for damage, but captures the anisotropic behavior, rate
475 dependence, tension-compression asymmetry and isotropic/kinematic hardening of the
476 material. 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 κ (see Appendix
480 for the definition), stress triaxiality ratio χ 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.

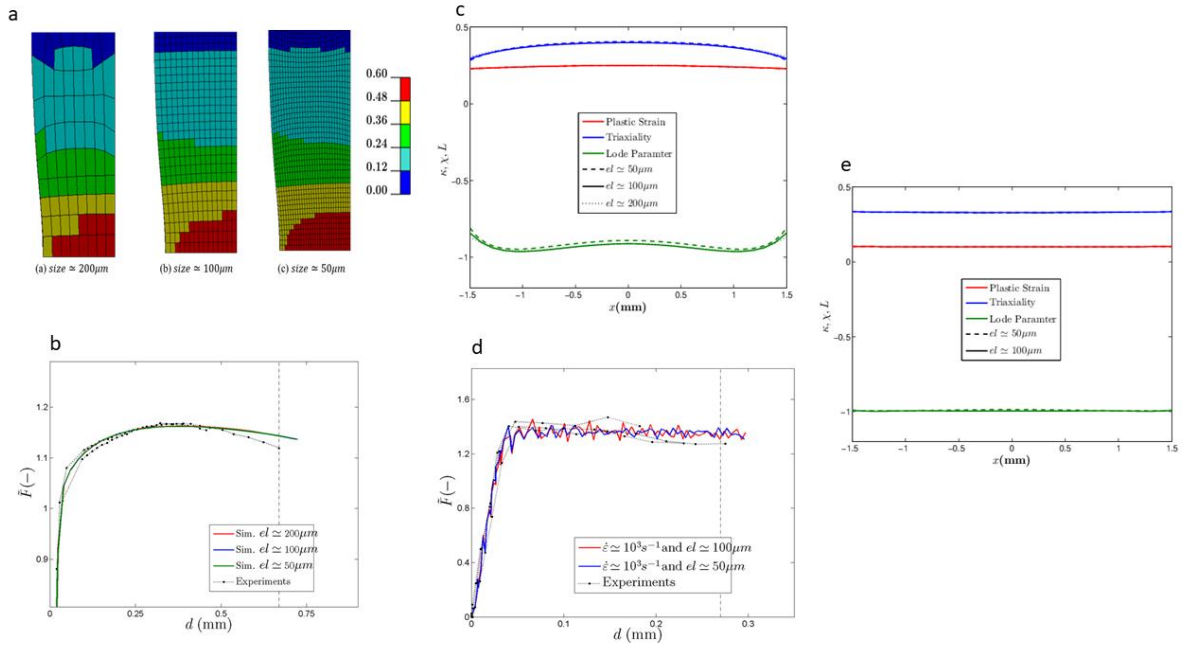


500 Figure 19: Geometries and boundary conditions for the numerical simulations. The green
 501 areas represent the imposed displacement (or fixture) and the red zones the symmetry
 502 conditions. The parts where internal variables are extracted are indicated by blue lines.
 503

504
 505 **3.1 Uniaxial tension**

506
 507 The smooth tension specimens were modelled with three mesh sizes of 50, 100 and 200 μm 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 μm 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

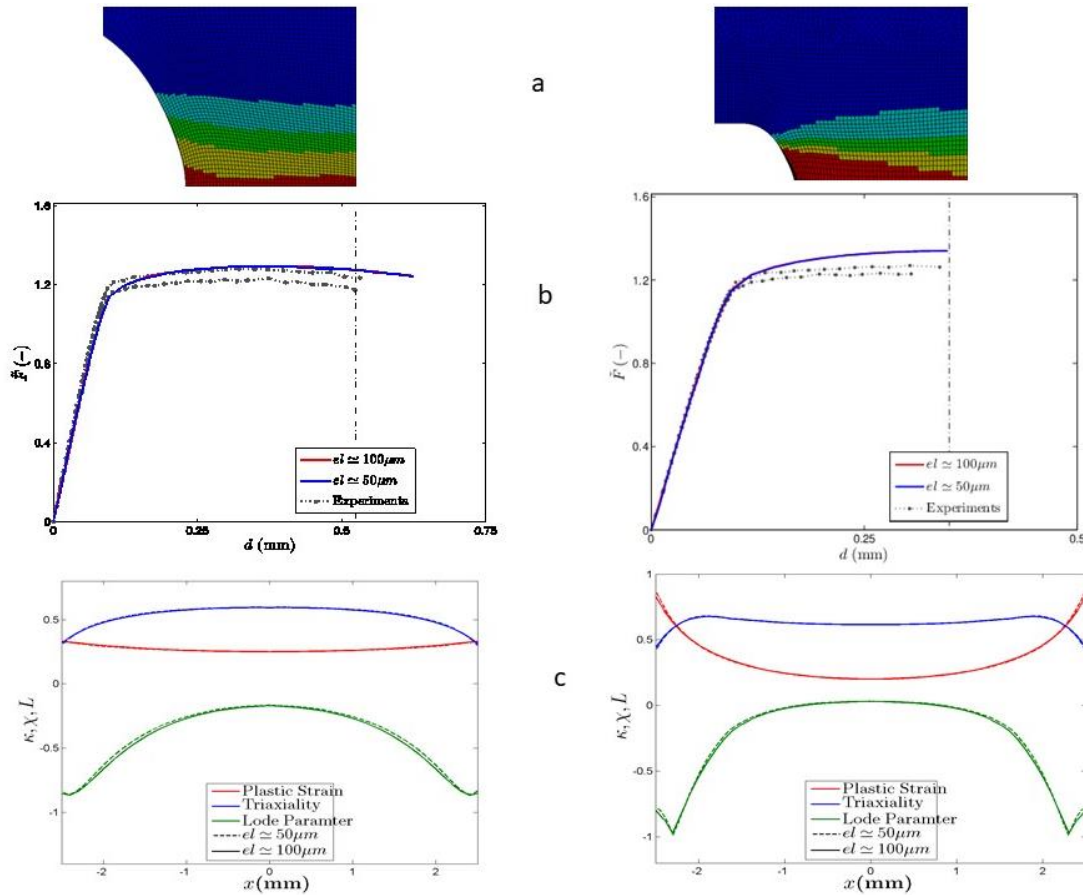


524
 525 Figure 20: Simulation of the tensile tests along TD on smooth specimens, using three different
 526 mesh sizes. a) comparison of the plastic strain fields at 10^{-3}s^{-1} , b) normalized load-
 527 displacement curves at 10^{-3}s^{-1} , c) transverse profiles of plastic strain κ , stress triaxiality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

532 In addition, the plastic strain κ , stress triaxiality χ 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 κ ,
 547 stress triaxiality χ 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 2mm and lower stress triaxiality χ), the maximum values of stress triaxiality are found at mid-
 554 width, and the stress triaxiality χ 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 χ). 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.

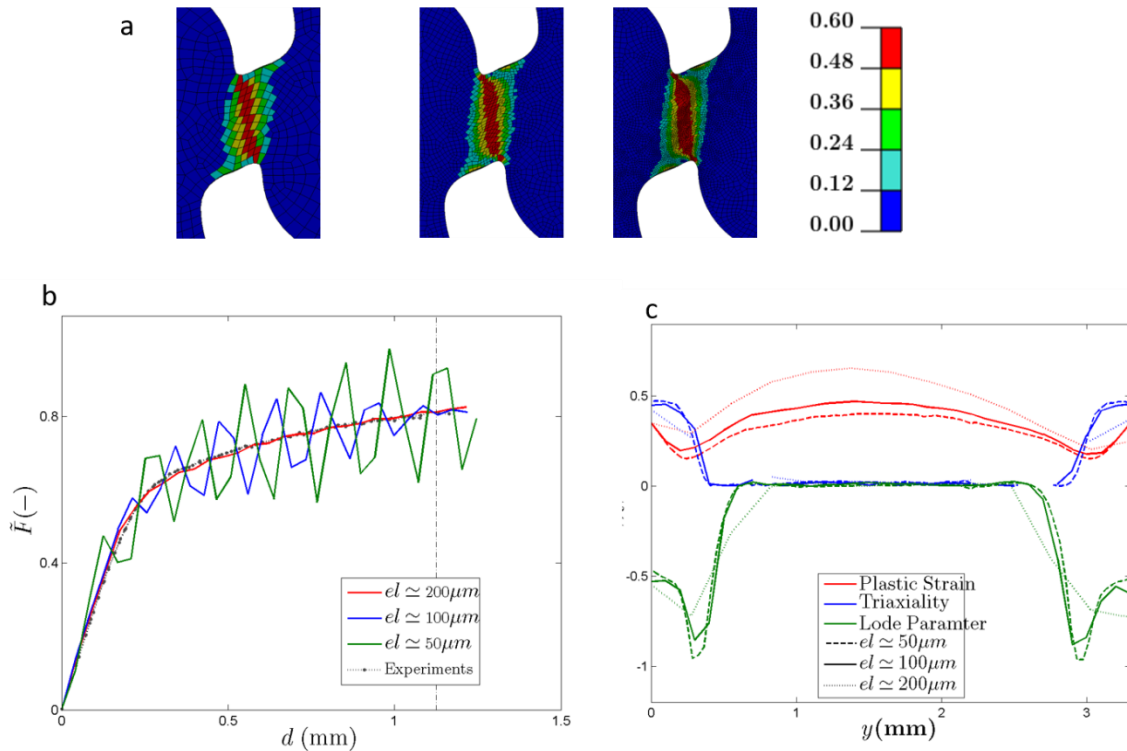
560
 561

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.

566

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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 κ , stress triaxiality χ 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 κ , the stress triaxiality ratio χ 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.

585

586

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589

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.

589

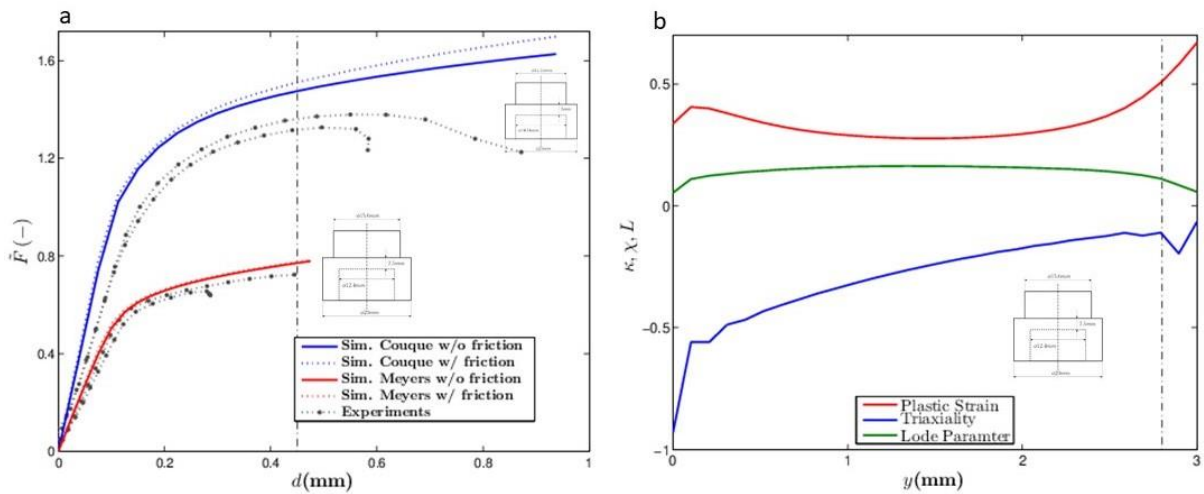
590

3.4. Compression-shear loading

591

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 ≈ -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 κ , stress triaxiality χ 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 κ , stress triaxiality χ 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

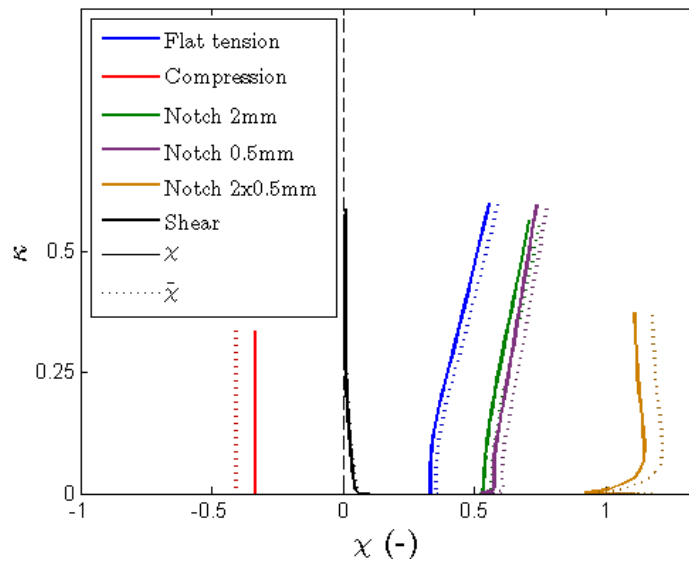
612

613 The local loading paths leading to fracture may be described through the evolutions of the stress
 614 triaxiality ratio χ 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 χ 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 κ as a function of the stress triaxiality ratio χ . $\dot{\epsilon} \approx 10^{-3} s^{-1}$, 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 χ ($\tilde{\chi} < \chi \approx -1/3$). 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 χ
 637 based on Von Mises stress is retained in the sequel.

638
 639 According to Figure 24, the stress triaxialities χ 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 κ , χ 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, κ	Stress triaxiality, χ	Lode parameter, L
RD	Smooth compression 10^{-3} s^{-1}	0.25	-0.33	1
	Shear	0.36	0.01	0
	Smooth tension 10^{-3} s^{-1}	0.2	0.38	-0.94
	Smooth tension 300 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} s^{-1}	0.16	-0.33	1
	Shear	0.39	0.01	0.01
	Smooth tension 10^{-3} s^{-1}	0.25	0.4	-0.9
	Smooth tension 300 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} s^{-1}	0.28	-0.33	1
	Shear	0.44	0.03	0
	Smooth tension 10^{-3} s^{-1}	0.44	0.49	-0.8
	Smooth tension 300 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} s^{-1} (approximations)	≈ 0.4	≈ -0.75	≈ 0.1
	Meyer 10^{-3} s^{-1}	0.67	-0.43	0.1
	Smooth compression 10^{-3} s^{-1}	0.38	-0.33	1
	Smooth tension 10^{-3} s^{-1}	0.15	0.36	-0.93

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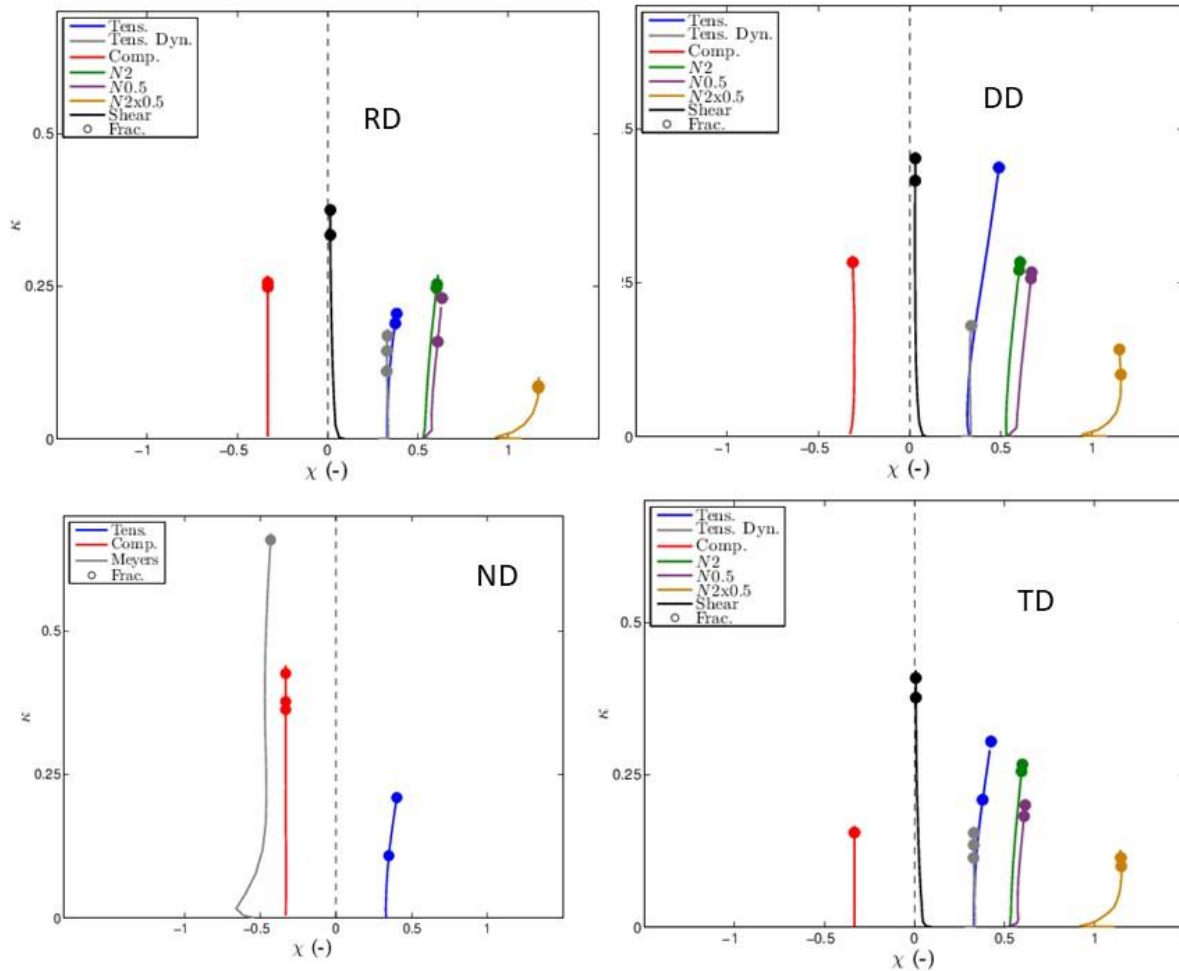


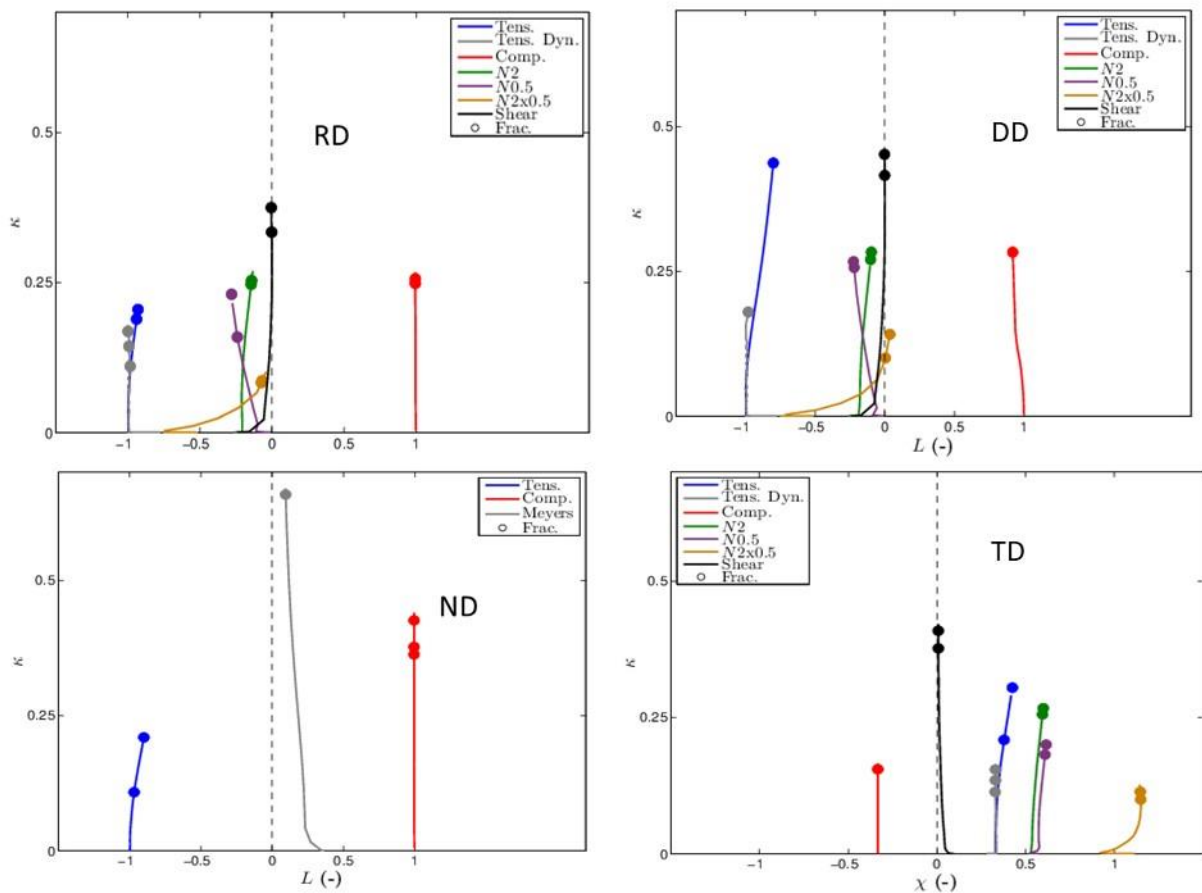
Figure 25: Plastic strain κ vs. stress triaxiality ratio χ for all loading directions. The dots indicate fracture.

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Figure 25 suggests a smaller strain to fracture when the triaxiality increases in absolute values, with a maximum ductility for χ 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 χ 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 χ 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 κ 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.

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682 Figure 26: Plastic strain κ 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 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 α/β 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**

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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:

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

736
737
738 where $(T, \underline{\underline{\varepsilon}}^e, \kappa, \underline{\underline{a}})$ represents temperature, elastic strain 2^{nd} order tensor, isotropic hardening
 739 variable (and cumulated plastic strain) and kinematic hardening variable 2^{nd} order tensor, and
 740 $(s, \underline{\underline{\sigma}}, r, \underline{\underline{X}})$ entropy, Cauchy stress 2^{nd} order tensor, isotropic hardening force and kinematic
 741 hardening force 2^{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 (α, K, C) represent the thermal dilatation coefficient, bulk modulus and kinematic

747 hardening related constant. The stiffness 4th 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 (λ, μ) 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 2nd order tensors, and (β, ρ, 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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