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# Investigation of residual stresses and modeling of tensile deformation in wire-arc additive manufactured 6061 aluminum alloy: Diffraction and elastoplastic self-consistent model

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1 **Investigation of residual stresses and modeling of tensile deformation in wire-**  
2 **arc additive manufactured 6061 aluminum alloy: diffraction and elastoplastic**  
3 **self-consistent model**

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15

16 **Abstract**

17 The aim of this work is to study the mechanical behavior of 6061 aluminum walls produced by  
18 Wire Arc Additive Manufacturing at the different scales of the material. The residual stresses of  
19 the parts are characterized on the surface and in the bulk thanks to X-ray and neutron diffraction  
20 techniques, respectively. Although the residual stress values determined in the close-surface are  
21 low (40 MPa for the maximum value), the mechanical state is different in depth with significant  
22 maximum residual stress levels (+/- 100 MPa). Tensile tests carried out on samples extracted from  
23 a WAAM wall in three orientations reveal that the mechanical properties of the manufactured  
24 material are slightly better compared to conventional material in T6 state. *In situ* X-ray diffraction  
25 experiments have also been performed under uniaxial tensile testing in order to determine strain  
26 pole figures. An elasto-plastic self-consistent model considering the microstructural characteristics  
27 and the main physical phenomena governing the material behavior, is proposed. The model agrees

28 with the experimental data and the simulations reproduce the main features observed at the scale  
29 of the diffracting volume.

30

### 31 **Keywords**

32 Strain pole figure, X-ray diffraction, Residual stresses, Wire and Arc Additive Manufacturing,  
33 Mechanical properties.

34

### 35 **Introduction**

36 In some cases, Additive Manufacturing (AM) allows to optimize the mechanical performance of a  
37 part while minimizing its environmental impact through reduced energetic and material use. Wire  
38 Arc Additive Manufacturing (WAAM) is an innovative concept derived from arc welding  
39 processes [1–3] that can be used for AM. It enables to improve the freedom of part design without  
40 being limited by a closed chamber. It is also one of the techniques with the highest deposition rates  
41 in AM of metallic materials leading to the capability to manufacture large meter-scale components  
42 at a lower cost [1,4]. Age hardened aluminum alloys such as aluminum 6061 (AA6061) are widely  
43 used in many structural applications due to some of their noteworthy properties such as high  
44 specific strength combined with a natural corrosion resistance. However, their well-known poor  
45 weldability [5,6] have made them very poor material candidates for AM processes. The recent  
46 development of a MIG (Metal Inert Gas) deposition process, based on a low-energy short-circuit  
47 transfer mode and called CMT (Cold Metal Transfer), have brought about a major change, making  
48 it possible to weld age hardened AA6061 [7–10]. However, those CMT-related studies have been  
49 limited to simple welding. In a very recent study [11], we have shown that it is possible to obtain

50 defect-free parts with the AA6061 using the WAAM process. In this study, wall samples were  
51 successfully built using two different deposition strategies. The obtained micro- and macro-  
52 structures have been found highly dependent on both process parameters and deposition strategy  
53 although both employed deposition strategies produced stacked structures composed of successive  
54 alternating layers of columnar and equiaxial grains. In addition, after heat treatment, a higher  
55 hardness was obtained for the built parts as compared to conventional forged products. In this  
56 previous study, the microstructure of AA6061 parts has then been analyzed in order to understand  
57 these hardening phenomena.

58 However, residual stresses developed during the WAAM building process have an obvious critical  
59 influence on the mechanical performance of the part, potentially leading to delamination from the  
60 supporting structure, part distortion or crack formation [12,13]. Those residual stresses arise from  
61 the high cooling rates, thermal gradients and volumetric changes due to phase transformations  
62 occurring over the building process and that are strongly related to the process parameters (heat  
63 input, welding speed, wire feed speed and welding trajectories) [14].

64 Amazingly, although aluminum is one of the most widely used alloys for AM techniques like  
65 WAAM [2,15,16], to date, no detailed studies on the mechanical behavior and the residual stress  
66 development are available, especially as regard to AA6061 [2,17]. Notwithstanding, the  
67 optimization of this technology necessarily requires an in-depth knowledge of the residual stress  
68 field being built in the parts as a function of the process parameters in order to ensure reliable  
69 performances. Moreover, in order to control the properties of the manufactured parts, it is also  
70 necessary to study in detail the influence of this process parameters on the mechanical properties  
71 at the different scales of the material.

72 The present work aims to study the influence of the WAAM manufacturing process and its related  
73 parameters on the mechanical properties.

74 First, the mechanical state of the parts has been characterized in the close-surface and in the bulk  
75 thanks to X-ray and neutron diffraction techniques, respectively. The residual stress analysis is  
76 actually a requirement to understand the mechanical behavior of the manufactured parts and to  
77 possibly optimize the process parameters and/or the post manufacturing heat treatment.

78 Secondly, tensile tests have subsequently been carried out to obtain the main features concerning  
79 the elastoplastic behavior of the material with respect to the process-induced microstructure.

80 Finally, a multiscale modeling of the elastoplastic behavior of the wire-arc additive manufactured  
81 6061 aluminum alloy is proposed. The modeling used in this work is based on a mean field  
82 approach (Elasto-Plastic Self-Consistent model, EPSC model) considering the microstructural  
83 characteristics and the main physical phenomena governing the material behavior. The purpose  
84 was hence to develop a tool to better understand the relationship between the mechanical properties  
85 and the microstructures produced by WAAM (grain sizes, crystallographic and morphological  
86 textures ...). This modeling was coupled with *in situ* X-ray diffraction tensile experiments.

87

## 88 **I. Experiment**

### 89 **I.1. Material and deposition process**

90 Since the deposition process has been extensively presented in a previous work [11], only a brief  
91 description will be done here. For this study, AA1050 aluminum alloy was used as substrate (plates  
92 with a size of  $2 \times 30 \times 250 \text{ mm}^3$ ) and a 1.2 mm diameter of AA6061 alloy wire was chosen as the  
93 filling material. The measured chemical composition of the welding wire is listed in Table 1.

94 AA6061 thin-wall samples with dimensions 150 (x) × 5 (y) × 55 (z) mm<sup>3</sup> have been deposited by  
 95 the MIG-CMT process developed by Fronius®. Each layer of these walls was deposited in a single  
 96 pass. Fore sake of clarity, three directions are set as follow: x, along the welding direction; y,  
 97 perpendicular to the travel direction of the welding torch and z, along the building direction. A  
 98 Cartesian robot with step-by-step motors equipped with a Fronius® TransPuls Synergic 3200 CMT  
 99 welding device composed the experimental set-up. The welding torch working distance was set  
 100 constant at 15 mm. A water-cooling system regulated the bottom of the clamped AA1050 baseplate  
 101 at 20 °C. Wire feed rate was set at 4.3 m.min<sup>-1</sup> and travel speed at 320 mm.min<sup>-1</sup>. A mixture of  
 102 argon and helium in a ratio of 80/20 has been used as shielding gas. Two deposition strategies (in  
 103 terms of torch trajectories sequence) were studied: unidirectional deposition (deposition without  
 104 change in the welding direction) and bidirectional deposition (with two alternating opposite  
 105 deposition directions).

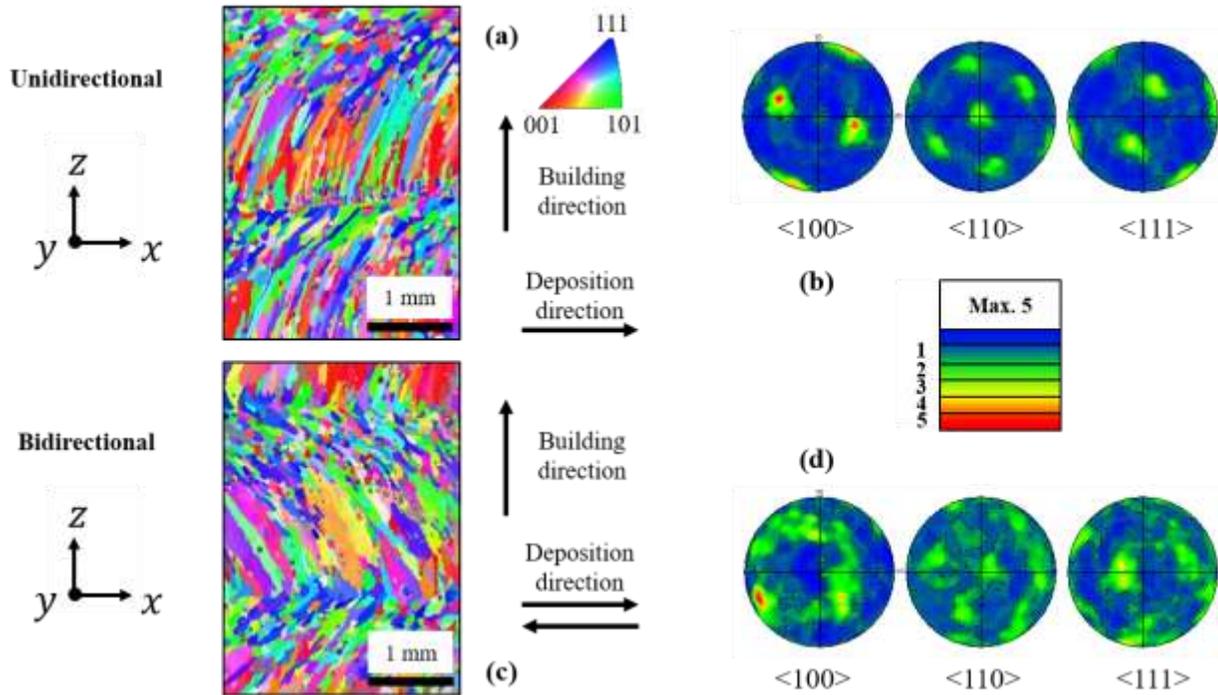
Elements	Mg	Si	Cr	Fe	Ti	Zn	Mn	Cu	Al
AA6061 wt%	0.9	1.0	0.1	0.3	-	0.2	0.1	0.2	Bal.

106 Table 1: chemical composition of the AA6061 welding wire measured using EDS.

107

108 The grain structures were analyzed by EBSD. Results are depicted in Figure 1 in the form of inverse  
 109 pole figure maps along the transverse direction y. The microstructure was examined in the middle  
 110 of the wall, in order to avoid the effects of arc initiation and extinction. Observations were also  
 111 made at different heights in the wall, as the microstructure of the first and last layers differs from  
 112 the remainder of the wall. The results presented here are based on mid-height observations.

113 Both unidirectional and bidirectional strategies induce the same structure of layers. This structure  
114 is composed of a stack of alternating equiaxed and columnar grains zones. Each deposited layer is  
115 made of columnar grains that epitaxially grow from the previously solidified grains present on the  
116 surface of the previous layer. A thinner zone of equiaxed grains is located at the top of each layer.  
117 Each new built layer partially melts the previous one. The growth direction of the columnar grains  
118 is related to the deposition direction (through the heat gradient that set during deposition). Thus,  
119 the bidirectional strategy, which alternates two opposite deposition directions, produces columnar  
120 grains growing along two alternating solidification directions, symmetrically distributed with  
121 respect to the transverse plane (y-z), while the unidirectional strategy maintains the same  
122 solidification direction layer after layer. The unidirectional strategy gives a single growth direction,  
123 resulting in a sharp  $\langle 100 \rangle$  texture component in the  $x + 65^\circ$  direction, as illustrated in the pole  
124 figures in Figure 1. The alternating strategy gives a more balanced texture with two reinforcements  
125 matching the two preferential solidification directions.



126  
 127 Figure 1: EBSD maps for WAAM samples built; (a) with the unidirectional strategy and (c) with  
 128 the bidirectional strategy. (b) and (d) show their respective pole figures.

129  
 130 For the residual stress determination, different walls have been built for this study. The analysis  
 131 was performed for the two deposition strategies and for two metallurgical states as follows:

- 132 • As-Deposited (denoted AD): material has simply been kept at -80 °C directly after printing in  
 133 order to prevent it from natural aging,
- 134 • T6 heat treatment (denoted T6): solid solutionizing at 530 °C during 1 h + quenching (water  
 135 cooling) + artificial aging at 175 °C during 8 h.

136 Concerning the *in situ* X-ray diffraction tensile experiments, the testing was performed with the  
 137 material in T6 condition built with the bidirectional strategy.

## 139 I.2. Subsurface residual stress analysis by X-ray diffraction

140 The average lattice strain within the diffracting volume  $V_d$  can be calculated from the shift of  
141 measured Bragg's angle,  $\theta(\varphi, \psi, hkl)$ , along the measurement direction defined by the angles  
142  $(\varphi, \psi)$  for a  $\{hkl\}$  reflection, as compared to its value in the stress-free material, as detailed in  
143 equation (1) [18]:

$$144 \langle \varepsilon(\varphi, \psi, hkl) \rangle_{V_d} = \ln \left( \frac{\sin \theta_0(hkl)}{\sin \theta(\varphi, \psi, hkl)} \right) \quad (1)$$

145  $\langle \rangle_{V_d}$  corresponds to an averaging operation over the diffracting volume  $V_d$ .  $\theta_0$  is the Bragg's angle  
146 of the stress-free material. In the case of a homogeneous and macroscopically elastically isotropic  
147 single-phase polycrystal, neglecting plastic deformation incompatibilities, the traditional equation  
148 for X-ray stress analysis is given with the  $\sin^2\psi$  method as:

$$149 \langle \varepsilon(hkl, \varphi, \psi) \rangle_{V_d} = \frac{1}{2} S_2(hkl) (\sigma_{11} \cos^2 \varphi + \sigma_{12} \sin 2\varphi + \sigma_{22} \sin^2 \varphi) \sin^2 \psi + \\ 150 \frac{1}{2} S_2(hkl) (\sigma_{13} \cos \varphi + \sigma_{23} \sin \varphi) \sin 2\psi + S_1(hkl) (\sigma_{11} + \sigma_{22}) \quad (2)$$

151 Where  $\frac{1}{2} S_2(hkl)$  and  $S_1(hkl)$  are the so-called X-ray Elastic Constants (XEC),  $\sigma_{ij}$  the components  
152 of the Cauchy's stress tensor in the sample orthonormal coordinates system (1, 2, 3). Regarding  
153 the XRD set-up used to perform the experiments, indexes 1, 2 and 3 respectively refer to the z, x  
154 and y directions of the part, defined ahead. It is usually assumed that, because XRD is a near-  
155 surface analysis, the diagonal component related to the normal to the analyzed surface direction -  
156 i.e.  $\sigma_{33}$  component - is equal to zero due to the low penetration depth ( $19.9 \mu\text{m}$  for  $\sin^2\psi = 0.3$   
157 with Co radiation). Note that with this assumption, information about the real stress-free lattice  
158 parameter  $\theta_0(hkl)$  is then no more necessary. In the present XRD analysis set-up,  $\sigma_{33}$  sample

159 coordinates matches  $\sigma_{yy}$  in the part i.e. stress related to the perpendicular to the travel direction of  
160 the welding torch.

161 Residual stress analyses were carried out with a four-circle Seifert XRD 3003 PTS diffractometer  
162 with a cobalt radiation. The X-ray beam output collimator was 1 mm in diameter. The diffraction  
163 peaks were recorded with a position sensitive detector. The determination of residual stresses was  
164 conducted using the  $\sin^2\psi$  method with the {331} diffraction peak at  $2\theta \approx 149^\circ$ . Diffractograms  
165 were recorded for sixteen different tilt angles  $\psi$  varying between  $-48^\circ$  and  $45^\circ$  and for azimuth  
166 angles  $\varphi$  of  $0^\circ$ ,  $45^\circ$  and  $90^\circ$ . The direction  $\varphi = 0^\circ$  corresponds to the welding direction. Peak  
167 positions were calculated using a pseudo-Voigt fitting function for the measured peak considering  
168 the dual  $K_{\alpha 1}$ - $K_{\alpha 2}$  contribution. Since the studied material displays large grains, oscillations of the  
169  $\psi$  angle ( $\Delta\psi : +/-5^\circ$ ) have been performed during the measurements so as to analyze a larger volume  
170 of material, optimizing the residual stress measurement conditions. The subsequent error induced  
171 has been demonstrated to be no more than 2 % on the stress values for the chosen oscillation  
172 magnitude [18,19]. The stress field investigation has been performed along a line following the  
173 building direction  $z$  in the middle of the sample in order to ensure analysis in the stationary regime.  
174 Ten different positions, distributed over the height of the wall, have been probed by XRD.

175

### 176 **I.3. Bulk residual stress analysis by neutron diffraction**

177 In order to determine the stresses in the bulk sample, additional Neutron Diffraction (ND)  
178 experiments have been carried out on the SALSA beamline [20] at the Institut Laue Langevin (ILL)  
179 in Grenoble (France). The gauge volume in the sample was set to  $2 \times 2 \times 2 \text{ mm}^3$  with a wavelength  
180 of  $1.62 \text{ \AA}$  as a compromise between grain size, mapping resolution and reasonable measurement

181 duration. The stress determination was performed by recording the {311} diffraction peak  
182 ( $2\theta \approx 88.5^\circ$ ).

183 The stress field investigation has been performed along a column set in the middle of the sample  
184 as for XRD investigations. Single peak fitting was performed using the LAMP software [21]  
185 enabling to obtain the peaks positions by fitting the experimental data with a pseudo-Voigt function  
186 and a polynomial background.

187 For each measurement point, lattice strains were determined along the three orthogonal directions  
188 x, y and z of the part to provide an evaluation of the three normal strain components  
189  $\langle \varepsilon(0,90, hkl) \rangle_{V_d} = \varepsilon_{xx}$ ,  $\langle \varepsilon(90,90, hkl) \rangle_{V_d} = \varepsilon_{yy}$  and  $\langle \varepsilon(0,0, hkl) \rangle_{V_d} = \varepsilon_{zz}$ .

190 With eq. (2), the normal stress components  $\sigma_{xx}$ ,  $\sigma_{yy}$ ,  $\sigma_{zz}$  are calculated [22].

191 Lattice strains within the diffracting volume are calculated from the Bragg's angle shift (Eq. (1)).  
192 In order to ensure reliable  $\theta_0$  measurements, especially regarding the possible evolution of the  
193 metallurgical state, the chemical composition and/or the microstructure during the building  
194 process, a specific procedure has been used, which is extensively detailed in [23]. The Bragg's  
195 angle of the stress-free material for each measurement point has been acquired thanks to  
196 measurements carried out over a series of mini-cubes of  $4 \times 4 \times 4 \text{ mm}^3$  cut thanks to electro-discharge  
197 from twin specimens (manufactured with the same process parameters) and matching the exact  
198 positions of the ND stress analyses of the walls. Cutting small coupon intends to relieve any  
199 macroscopic stress. In order to ensure that potential residual intergranular strains present in these  
200 mini-cubes do not interfere with the stress-free parameter determination, measurements have been  
201 achieved along numerous directions and then averaged.

202

#### 203 I.4. XEC determination method by XRD

204 The residual stress analysis through XRD or ND requires the accurate knowledge of the XEC,  
205  $\frac{1}{2} S_2(hkl)$  and  $S_1(hkl)$ . The XEC values for aluminum alloys can greatly vary according to the  
206 considered type of alloy, the quantity of alloying elements and the precipitation state [24,25]. No  
207 experimental value is actually available in the literature for the one studied in this work. It is  
208 therefore relevant to experimentally determine the XEC in order to consider the precipitation state  
209 and the microstructural parameters inherent to the studied parts.

210 These XEC can be determined when a sample is elastically loaded *in situ* in a diffractometer and  
211 the corresponding lattice strains are measured for different sample orientations and applied loads  
212 [18,26]. In this method, a quasi-isotropic single-phase material having no preferential  
213 crystallographic orientation (i.e. elastically anisotropic grains in a randomly distributed  
214 crystallographic orientation sample) is then assumed.

215 A known macroscopic stress  $\sigma^A$  is applied to the studied sample and the lattice strain  
216  $\langle \varepsilon(\varphi, \psi, hkl) \rangle_{V_d}$  is measured as a function of this mechanical stress. If a uniaxial tensile test is  
217 performed  $\sigma_{11}^A$ , the relation between the lattice strain and the average stress is:

218

219 Where 1 refers to longitudinal direction of the dogbone-shaped tensile specimen.

$$220 \langle \varepsilon(\varphi, \psi, hkl) \rangle_{V_d} = \frac{1}{2} S_2(hkl) [\sigma_{11}^A \cos^2 \varphi] \sin^2 \psi + S_1(hkl) [\sigma_{11}^A] \quad (3)$$

221 At  $\varphi = 0^\circ$ , we finally obtain:

222

$$223 \langle \varepsilon(\varphi, \psi, hkl) \rangle_{V_d} = \frac{1}{2} S_2(hkl) \sigma_{11}^A \sin^2 \psi + S_1(hkl) \sigma_{11}^A \quad (4)$$

224

225  $\langle \varepsilon(0, \psi, hkl) \rangle_{V_d}$  is measured for different values of the applied stress ( $\sigma_{11}^A$ ) and different values  
226 of declination angle ( $\psi$ ). The equation (4) can be used to fit XECs when  $\sigma_{11}^A$  is known. The  
227 measured  $\langle \varepsilon(0, \psi, hkl) \rangle_{V_d}$  values can be plotted according to  $\sin^2 \psi$  for different applied loads. If  
228 the texture influence is negligible, these curves can be fitted by linear functions. The slopes (resp.  
229 the intercepts) of these fits can be plotted versus the applied stress  $\sigma_{11}^A$ , and then fitted with a linear  
230 function where the resulting slope is the  $\frac{1}{2} S_2(hkl)$  XEC (resp.  $S_1(hkl)$ ).

231 Flat dog-bone tensile specimens with a strain gauge section of  $13.65 \times 8 \times 1 \text{ mm}^3$  were cut out by  
232 electro-discharge machining from the thin-walls samples considering three different directions  
233 with regard to the Welding Direction (WD): parallel (denoted WD), perpendicular (denoted BD  
234 for Building Direction) and inclined at  $-65^\circ$  from the deposition direction (denoted GD for Grain  
235 Direction). For the last tensile sample, the tensile direction is parallel with the long axis of the  
236 grains (see Fig. 1). *In situ* tensile tests were performed at room temperature within a four-circle  
237 Seifert XRD 3003 PTS diffractometer with a Co- $K_\alpha$  radiation and using a Deben tensile device  
238 adaptable to the Eulerian cradle of the goniometer. *In situ* lattice strain measurements, performed  
239 on  $\{331\}$  diffraction peak, were conducted in load-control mode with a load cell of 5000 N capacity  
240 and 0.1 N resolution. The  $\langle \varepsilon(\varphi, \psi, hkl) \rangle_{V_d}$  lattice strains were calculated with equation (1). In this  
241 case,  $\theta_0$  is the initial Bragg's angle of the sample at 0 MPa - i.e. when no load is applied. This  
242 reference angle has been measured in the undeformed samples for each probed direction  $(0, \psi)$ :  
243  $\theta_0(0, \psi, 331)$ . The measured lattice strains are therefore elastic strains induced by the mechanical  
244 loading and do not consider the potential contributions from residual strains due to the process. It  
245 should be however noticed that although residual stresses (see section III.2) exist in the built thin-  
246 wall, tensile sample machining actually removes the constraint provided by the remainder of the  
247 wall so that the macroscopic stresses will be mostly relaxed in the studied samples. The

248 experimental procedure consists in loading the sample at various load magnitudes, in the elastic  
249 regime: 0, 20, 40, 60, 80, 100, 120, 140, 160, 180 MPa and for several specimen orientations:  $\varphi =$   
250  $0^\circ$  and  $\psi = \{0^\circ; 6^\circ; 12^\circ; 18^\circ; 24^\circ; 30^\circ; 36^\circ; 42^\circ, 48^\circ\}$ .

### 251 **I.5. *In situ* measured strain pole figures**

252 *In situ* XRD analysis have been achieved under uniaxial tensile testing in order to determine strain  
253 pole figures in elasto-plasticity for the three tensile specimens (WD, BD and GD) and to quantify  
254 the influence of microstructural parameters (texture and grain shape) on the global mechanical  
255 behavior. The same experimental setup presented in the previous section has been used. During the  
256 *in situ* diffraction experiments, data were collected at 5 given macroscopic strain levels (0, 0.2, 0.5,  
257 0.8 and 3 %) for several specimen orientations. The samples were speckled, and images were  
258 recorded during the test. Digital image correlation was then processed to obtain the macroscopic  
259 total strain fields.

260 Incomplete strain pole figures were measured with tilt and azimuth angles ranging from  $0^\circ$  to  $50^\circ$   
261 and from  $0^\circ$  to  $90^\circ$ , respectively, at each chosen total strain state. diffractograms were recorded  
262 along different  $(\varphi, \psi)$  directions with a step of  $5^\circ$  for both  $\psi$  and  $\varphi$  angles. The  $\varphi$  upper value was  
263 limited to  $50^\circ$  owing to the shadowing of the tensile device. In order to determine the corresponding  
264 incomplete strain pole figure, a total of 209 diffractograms were recorded for each deformation  
265 step. For these investigations, at  $\psi = 90^\circ$  and  $\varphi = 90^\circ$ , the scattering vector  $\mathbf{Q}$  was aligned with  
266 the Longitudinal Direction (denoted LD – i.e. the loading direction).  $\mathbf{Q}$  was parallel to the  
267 Transverse Direction (denoted TD) at  $\psi = 90^\circ$  and  $\varphi = 0^\circ$ . The lattice strains  $\langle \varepsilon(\varphi, \psi, hkl) \rangle_{V_d}$   
268 were calculated with equation (1). The reference Bragg's angle has been measured in the  
269 undeformed samples for the {331} diffracting peak and for each  $(\varphi, \psi)$  direction:  $\theta_0(\varphi, \psi, 331)$ .

270

## 271 II. Model

272 A mean field approach, the self-consistent model, has been used to describe the mechanical  
273 behavior and to interpret the collected diffraction data. The objective is to replace the real  
274 heterogeneous material with a fictitious homogeneous material which has the same behavior at the  
275 macroscopic scale by using a homogenization procedure. In the present work, the model developed  
276 in [27] is used and will not be described in details here. For FCC metals, the plastic deformation  
277 was simulated assuming  $\{110\}\langle 111 \rangle$  slip systems. Each grain is supposed to be an ellipsoidal  
278 elastoplastic inclusion, embedded in an elastoplastic Equivalent Homogeneous Medium (EHM)  
279 representing the average properties of all the grains in the polycrystal. The EPSC model has the  
280 capability to account for the crystallographic and morphological (grain-shape) textures of the  
281 material [28]. Each grain is described by an elastic constant tensor, crystallographic slip directions  
282 and planes in order to predict the strain/stress state of the grain under a mechanical load. A  
283 hardening matrix,  $H^{gh}$ , is introduced to describe the CRSS (Critical Resolved Shear Stress)  
284 changes with strain for each deformation mode  $g$ , as a function of the plastic slip on the other  
285 deformation systems  $h$  for each grain:

$$286 \quad \dot{\tau}_c^g = \sum_h H^{gh} \dot{\gamma}^h \quad (4)$$

287  $\dot{\gamma}^h$  is the plastic slip rate on the  $h$ -th active system.  $H^{gh}$  are the hardening coefficients that  
288 characterize self- (with  $h = g$ ) and latent- (with  $h \neq g$ ) hardening between deformation systems  
289  $g$  and  $h$ .

290 The latent hardening coefficients are assumed to be described as follow [29]:

$$291 \quad H^{gh} = qH^{gg} \quad (g \neq h) \quad (5)$$

292 The  $q$  factor defines the degree of latent hardening. This factor is chosen in order to provide the  
293 best accordance between the experimental and simulated and stress-strain curves in plasticity.

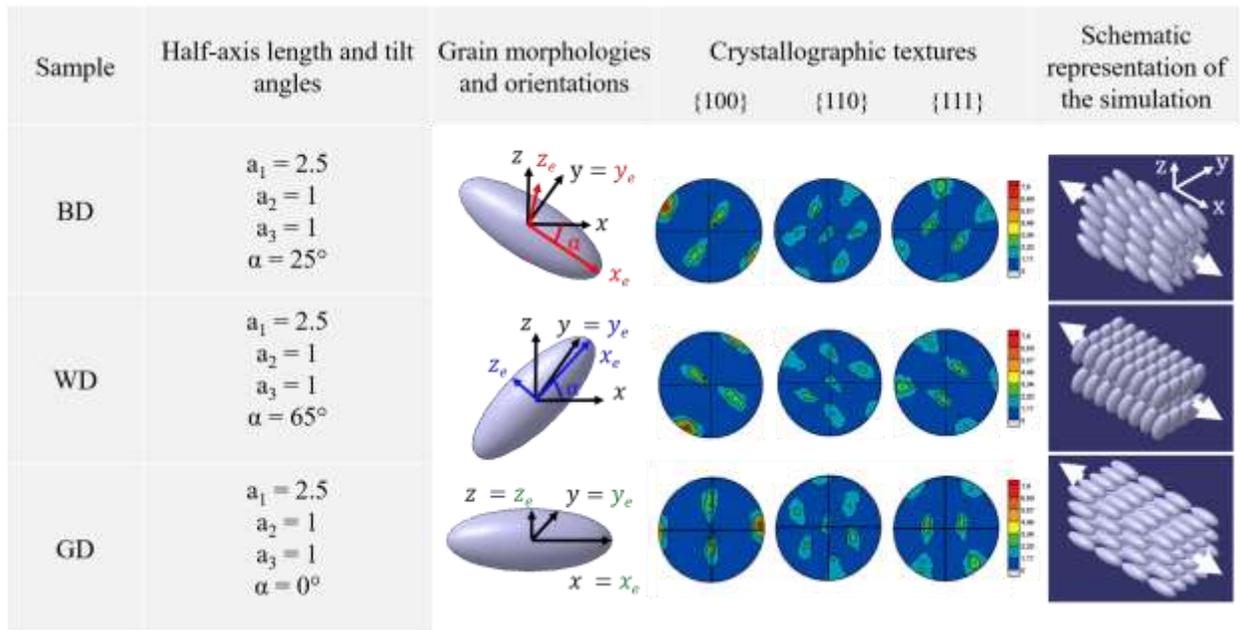
294  
295 Due to the lack of data for AA6061 alloy, the elastic constants of aluminum single-crystal used in  
296 the simulation are assumed to be those of pure aluminum [18]:  $c_{11} = 108.2$ ,  $c_{12} = 60$  and  
297  $c_{44} = 28.5$  MPa.

298 The CRSS value was chosen in order to reproduce the elastoplastic transition observed on the  
299 macroscopic stress – strain curve for the sample WD. The same value was used to simulate the  
300 behavior along the two other loading directions.

301 The EPSC model has the capability to account for the grain-shape and its evolution. Based on  
302 EBSD results, the simulation considers that the shape of the inclusion (grain) is ellipsoidal with  
303 half axes defined by  $a$ ,  $b$  and  $c$  so that  $a = 2.5 \neq b = c = 1$ . The initial texture is represented  
304 by a discrete distribution of 2000 ellipsoidal grains with Euler angle orientations weighted to  
305 reproduce the crystallographic texture shown in Figure 1. The grain orientation and  
306 crystallographic texture used to simulate the behavior of the WD, BD and GD samples are detailed  
307 in Figure 2. In the case of the WD sample, the inclusion long-axis  $x_e$  of the ellipsoid coordinates  
308 system  $(x_e, y_e, z_e)$  is tilted by an angle  $\alpha = 65^\circ$  from the  $y$ -axis of the part coordinates system  $(x, y,$   
309  $z)$ . In the case of the BD sample,  $\alpha = 25^\circ$ . For the GD sample, the ellipsoid coordinates system  
310 coincides with the one of the part.

311

312



313  
314 Figure 2: grain orientation and crystallographic texture used to simulate the behavior of the WD,  
315 GD and BD samples.

316  
317 The set of material parameter values giving the best agreement between the simulated and the  
318 measurements are presented in Table 2.

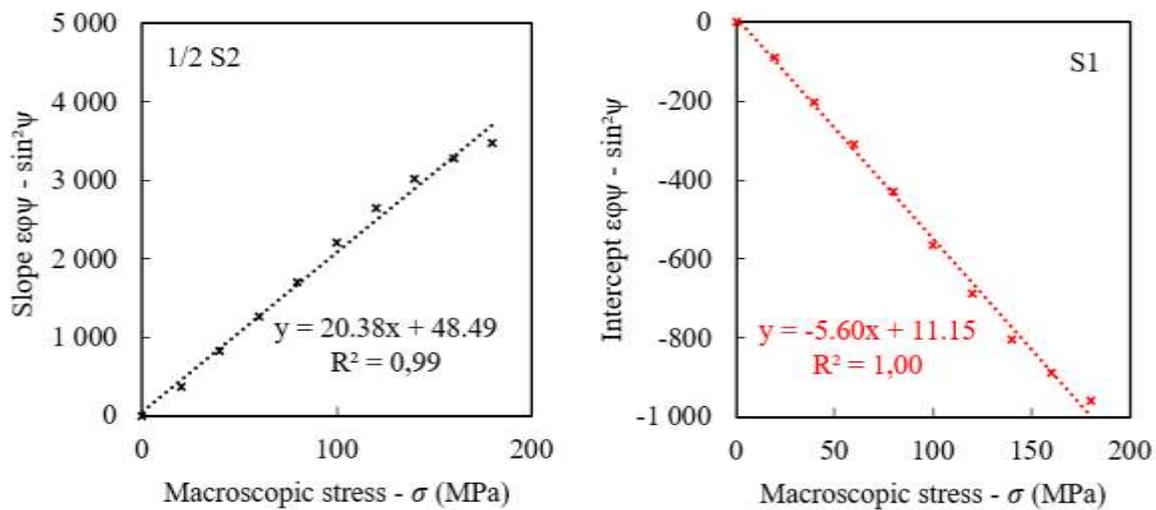
$H^{gg}$ (MPa)	$\tau_c^g$ (MPa)	$q$
30	115	1.1

319  
320 Tables 2: material parameters used in the simulations.

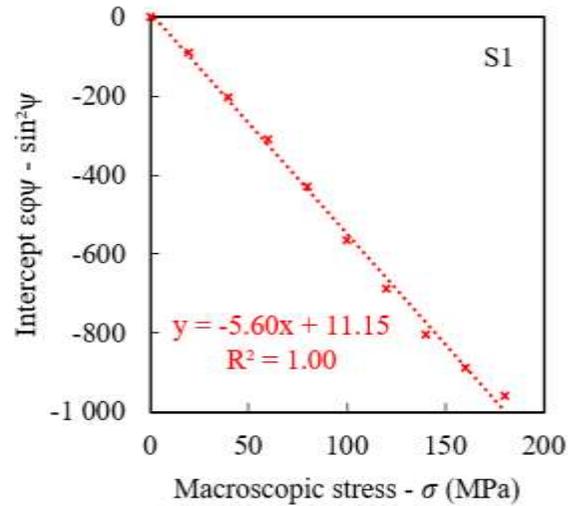
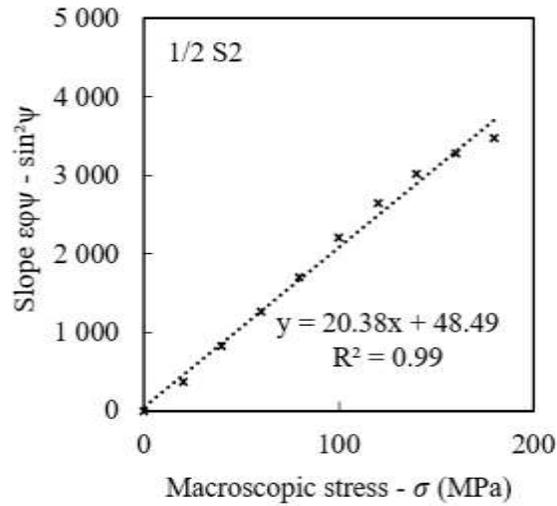
321  
322 **III. Results and discussion**

323 **III.1. XEC**

324 Based on the procedure described in section I.4 and using Eq. (4), XECs can be estimated by  
325 plotting the slope (Fig. 3.a) and intercept (Fig. 3.b) of the linear regression versus the applied  
326 macroscopic stress. The resulting data show a clear linear dependence with the applied macroscopic  
327 stress. According to the explanations given in section I.4, the slope of these linear functions  
328 corresponds respectively to  $\frac{1}{2}S_2(hkl)$  and  $S_1(hkl)$ . The correlation coefficients,  $R^2$ , are very close  
329 to 1 and, *a posteriori*, demonstrate the validity of assuming an isotropic macroscopic mechanical  
330 behavior for the material.



331



(a)

(b)

Figure 3: line fitting data from the procedure described in section I.4: a) slope and b) intercept versus the applied stress.

These values were obtained for the 3 specimens collected in the three directions WD, BD and GD.

The results are summarized in Table 3. They do not show significant differences.

The XECs,  $\frac{1}{2} S2\{hkl\}$  and  $S1\{hkl\}$ , can also be calculated by modeling the behavior of the polycrystalline aggregate [27,30]. Therefore, these constants were also calculated using an elastic self-consistent model [27] with the elastic constants of the pure aluminum single-crystal ( $c_{11} = 108.2$  GPa,  $c_{12} = 60$  GPa, and  $c_{44} = 28.5$  GPa [18], commonly used in the literature). These values can also be compared with those obtained in the literature for other aluminum alloys (see Table 3).

Tensile sample	$\frac{1}{2}S_2(hkl) (\times 10^{-6} \text{ MPa}^{-1})$	$S_1(hkl) (\times 10^{-6} \text{ MPa}^{-1})$
BD	20.4	-5.6
GD	22.4	-6.2
WD	21.3	-5.6
Mean value	21.3 +/- -1.0	-5.8 +/- -0.4
Elastic self-consistent simulation	19.2	-4.9
Pure Al [18]	18.9	-4.9
Al-2014-T6 [31]	16.8	-
Al-2024-T351 [31]	18.6	-
Al-7075 T6 [31]	17.5	-

346 Table 3: comparison of the experimental XECs with values from the literature and the values  
347 obtained with the elastic self-consistent model.

348

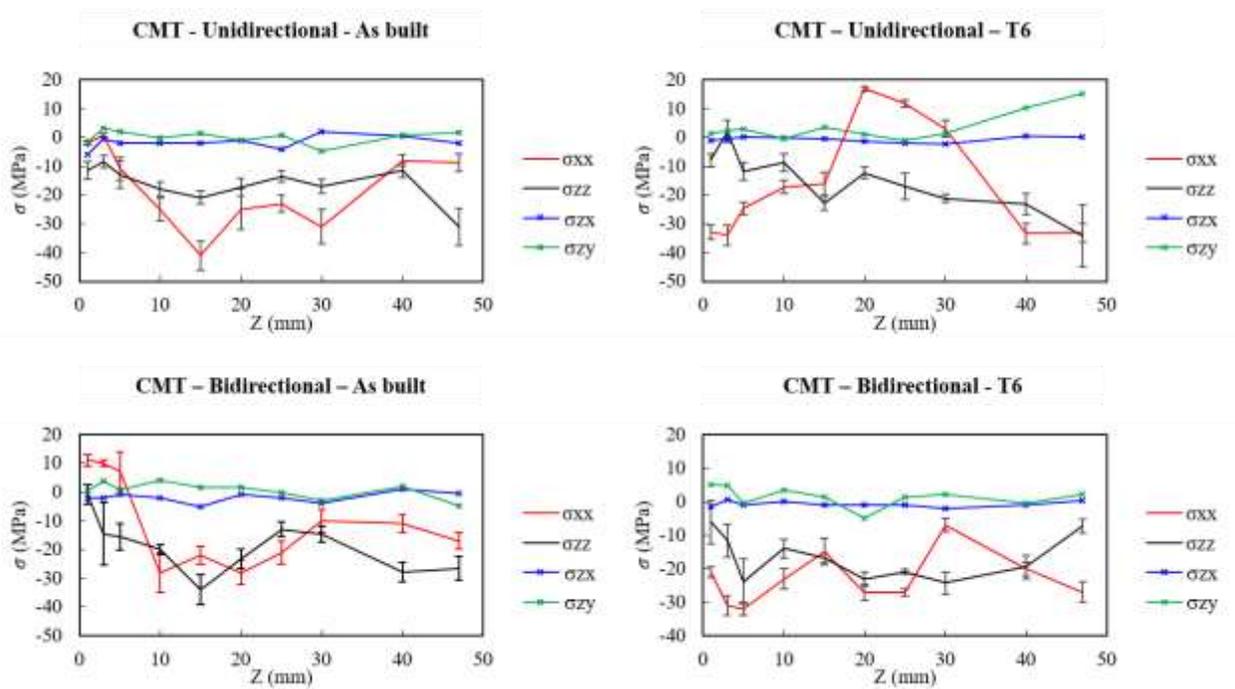
349 It can be noted that the experimental values are relatively different (about 10 %) from those  
350 obtained by modeling or those gathered from the literature for pure aluminum. The alloying  
351 elements present in the material as well as the distortion of the matrix around the precipitates can  
352 modify the XEC values [24,25]. This difference between pure aluminum and its alloys seems to be  
353 confirmed by the experimental values available for other types of aluminum alloys (Table 3). In  
354 the light of the weak differences observed in the three directions as well as the significant difference  
355 between the experimental values and the available data (calculation and literature) for pure  
356 aluminum, we thus made the choice to use the average values obtained experimentally for the stress  
357 analysis by XRD:  $\frac{1}{2}S_2(331) = 21.3 \cdot 10^{-6} \text{ MPa}^{-1}$  and  $S_1(331) = -5.8 \cdot 10^{-6} \text{ MPa}^{-1}$ . Note that the  
358 same XEC values have been used for the data analysis obtained by ND due to the quasi macroscopic  
359 elastic isotropy of the studied material.

360

361

362 **III.2. Residual stresses**

363 The stress profiles determined on the surface for each measured point distributed over the height z  
364 of the walls are shown in Figure 4. The position z = 0 mm corresponds to the top surface of the  
365 substrate. As a reminder, the  $\sigma_{33}$  stress in the direction normal to the probed surface is assumed to  
366 be null in the case of XRD analysis. This stress component actually matches  $\sigma_{yy}$  stress in the part  
367 coordinates system - i.e. along the transverse direction of the wall.



368  
369 Figure 4: residual stress analysis by XRD along the welding (WD, x-direction) and building (BD,  
370 z-direction) directions versus the distance from top surface of baseplate.

371 Whatever the printed sample analyzed, very low and similar levels of residual stresses are found.  
372 It seems that either the fabrication process completely relaxes the stresses or edge effects are  
373 responsible for the removal of residual stresses from the surface. The analyses have been conducted  
374 on surfaces in direct contact with the surrounding atmosphere. The stress level is globally

375 equivalent throughout the height of the different walls. Although residual stress values are low in  
376 the probed close-surface volume (with a maximum of 23 MPa for the average values and 40 MPa  
377 for the maximum value), the mechanical state might be different for deeper probed volumes.

378 ND experiments are thus relevant since it provides additional information leading to a better  
379 understanding of the mechanical state of the parts produced by WAAM. Significant maximum  
380 residual stress levels, more than +/- 100 MPa, are observed in the three measurement directions as  
381 observed in Figure 5. The maxima observed are as follow:

$$382 \quad \sigma_{xx} = +/- 119 \text{ MPa}, \sigma_{yy} = +/- 105 \text{ MPa}, \sigma_{zz} = +/- 110 \text{ MPa}.$$

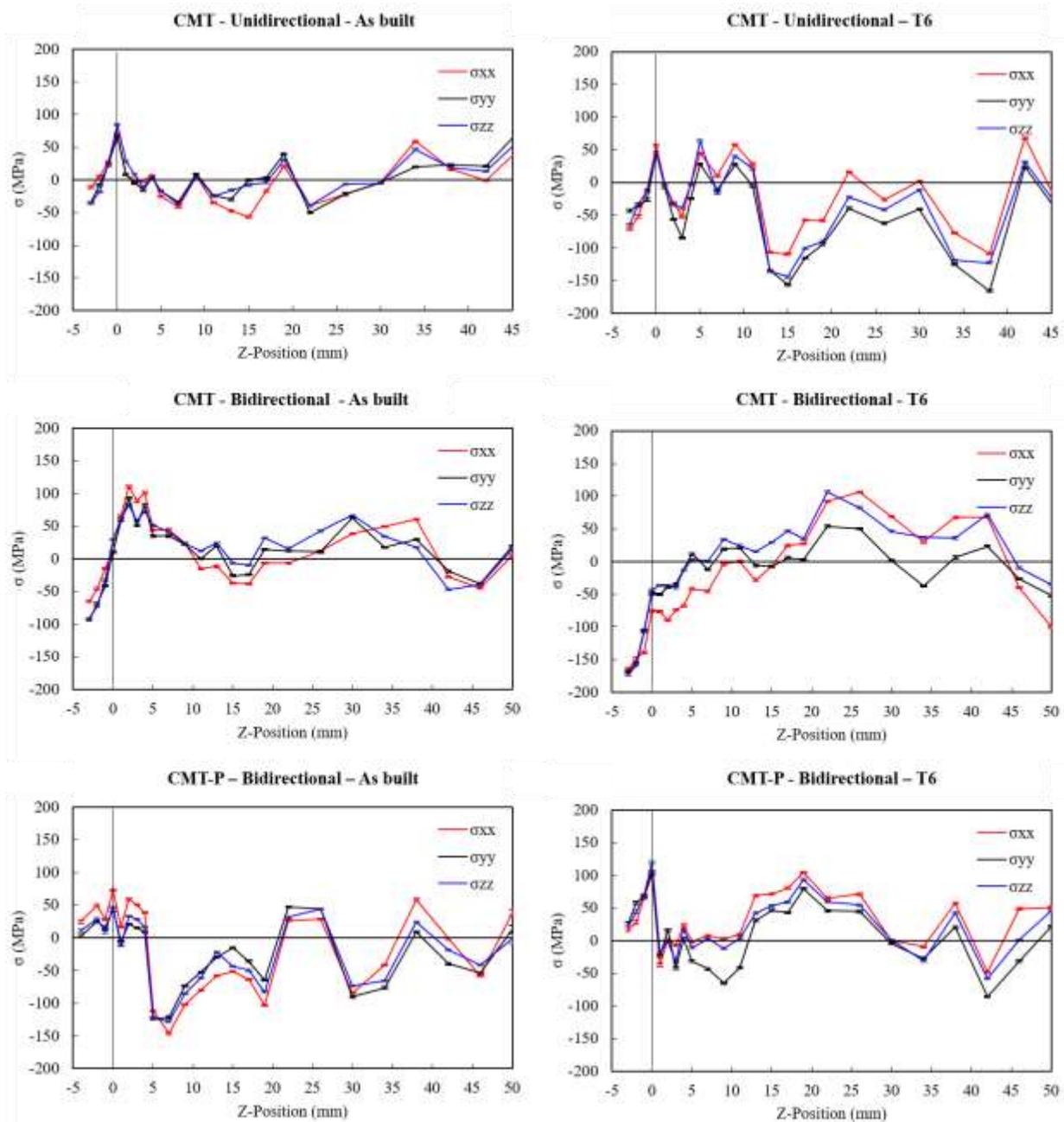
383 Slightly higher residual stress values were observed along the welding direction (x). These results  
384 are in agreement with the literature for other alloys [13,14,32]. However, the measured values in  
385 the two other directions are quite close, in contrast to the values observed in these previous works.  
386 Unidirectional and bidirectional strategies give similar stress levels and the same type of variations  
387 along the three main directions x, y and z. The same trend is observed for the T6 heat treatment  
388 with a modification of the stress profile.

389 In this study, the effect of arc mode type in cold metal transfer (CMT) process on the residual stress  
390 evolution has been also studied. More specifically, the cold metal transfer pulse (CMT-P) process  
391 [33] which is a combination of traditional CMT welding process and pulse welding process, was  
392 employed and its influence on the residual stress was analyzed. AA6061 thin-wall samples with  
393 the same dimensions (i.e. 150 (x) × 5 (y) × 55 (z) mm<sup>3</sup>) have been deposited using the MIG-CMT-  
394 P process with the bidirectional strategy. Figure 5 compares the residual stress distribution obtained  
395 with CMT-P and CMT processes.

396 In the case of CMT-P process, we find the trends observed by different authors [13,14,32] for other  
397 alloys. When the part has just been deposited by WAAM and is still clamped to the substrate, the  
398 stresses are generally more tensile in the wall and compressive in the substrate. During unclamping,  
399 more freedom is brought to the system baseplate + wall, with a reorganization of the mechanical  
400 fields. The part tends to contract at the top and expand near the baseplate. This is why distortions  
401 are observed. These distortions are accompanied by a redistribution of internal stresses. The  
402 stresses decrease with the height of the wall and can even reach compression at the top of the wall.  
403 In the baseplate, the stress evolves in the opposite way until it reaches tension in the lower part. In  
404 the transition zone between the baseplate and the first deposited layers, the tensile stresses reach a  
405 maximum.

406 As shown in Figure 5, in the case of CMT process, the observed trends are different. The residual  
407 stress values are negative (compression at  $Z = -4$  mm) within the baseplate. The stress profile  
408 evolves rapidly from the first layers towards tensile stress values. A maximum is reached in the  
409 first layers of the wall (close to the baseplate). A second tensile maximum is observed in the last  
410 layers ( $Z = 40$  mm) and the stresses remain relatively low between these two points. In the case of  
411 CMT process, the sample shows a stress profile similar to a clamped sample as described in the  
412 literature (compressive stress in the substrate and tensile stress in the wall [14]). It is possible that  
413 the clamping conditions or the deposition setup have allowed the substrate more freedom to deform  
414 during deposition, leading to a different stress distribution.

415



416

417 Figure 5: residual stress distribution obtained by ND with CMT and CMT-P processes for

418 unidirectional and bidirectional strategies.

419 To explain the difference between residual stress values in the probed close-surface volume and

420 the mechanical state for deeper probed volumes, it should be recall that the residual stresses

421 determined by XRD only concern the near-surface zone of the material and vary between -40MPa  
422 and 10MPa with a predominantly compressive state. These stresses are low, whereas the stresses  
423 determined by ND in the core sample show larger variations (between -150MPa and 100MPa) and  
424 mechanical states that are locally compressive which turn into tension in other regions. If we  
425 consider the walls (without considering the substrate) in the AD material, the range is smaller (-50  
426 MPa, 100 MPa) but remains larger than the values obtained at the surface. Internal stresses develop  
427 in built components due to the high cooling rates, the thermal gradients and the volumetric changes  
428 arising during phase transformations occurring during the process. Thermal gradients tend to  
429 induce tensile stresses in the last deposited layer. The residual stresses decrease and could thus  
430 even reach negative values in the underlying layers [34]. The highest stress values are generally  
431 observed near the substrate. This is consistent with the results observed for the AD material with  
432 the unidirectional strategy and the CMT process. The effect of microstructure is more difficult to  
433 understand, and the material used here is particularly sensitive to this effect since it is a  
434 precipitation-hardened aluminum alloy. T6 treatment, which significantly modifies the  
435 microstructure, changes the stress profile, as shown by the results obtained in Figure 5 (CMT-P  
436 Bidirectional, CMT Bidirectional and CMT Unidirectional) before and after treatment. As surface  
437 cooling conditions are specific, it is logical to determine residual surface stresses (XRD) different  
438 from the stresses obtained in the bulk (ND).

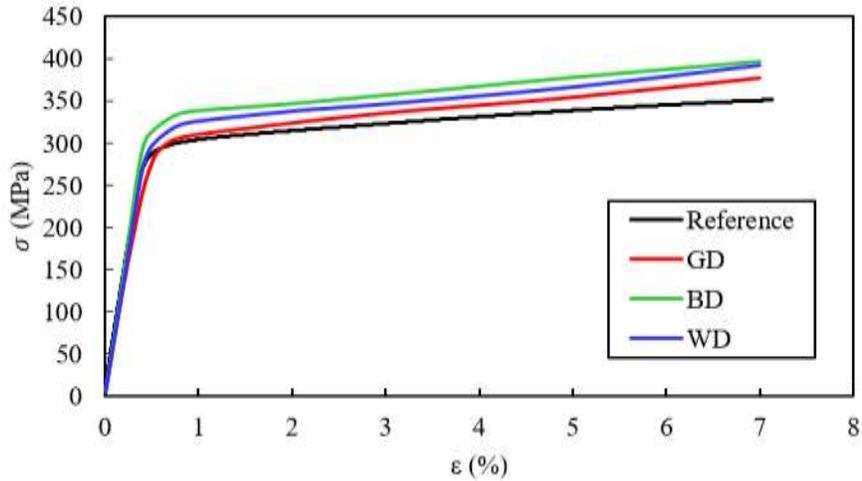
439

### 440 **III.3. Mechanical properties**

441 The results obtained at the macroscopic scale with the *in situ* tensile tests allowed to achieve the  
442 tensile curves. It should be noted that the maximum imposed strain is 7 % due to a systematic  
443 unexpected stop of the tensile machine after a very large number of measurements during the XRD

444 tests. The results of the tensile tests for the 3 three specimen orientations (WD, BD, GD) have been  
445 compared to a reference sample extracted from a rolled plate in T6 condition and are presented in  
446 Figure 6.

447



448

449 Figure 6: stress-strain curves for WD, BD and GD tensile specimens compared to the reference  
450 sample extracted from a rolled plate in T6 condition.

451

452 The tensile properties of additively manufactured specimens are similar or slightly better than those  
453 obtained for the reference sample obtained from a rolled 6061-T6 plate. The related mechanical  
454 parameters are summarized in Table 4. The maximum strength  $R_{ms}$  recorded at 7 % macroscopic  
455 strain is systematically higher than the one observed for the reference specimen. However, no  
456 significant difference between the different directions is observed: the presence of equiaxed grains  
457 has probably enabled to limit the effect of the columnar grains. It should be noted that the properties  
458 of the GD specimen (with the loading direction parallel to the grain growth direction) only show  
459 slightly lower properties than the others (especially as regard to the yield strength,  $R_e$ ). A similar  
460 study (results not shown here for the sake of clarity) has been performed with the second deposition

461 strategy. It has revealed that the two strategies display equivalent mechanical properties, certainly  
462 due to the presence of these equiaxed grains. The crystallographic texture for the bidirectional  
463 strategy, which is more homogeneous at the part scale due to the two growth directions (Fig. 1),  
464 does not seem to have a significant effect on the mechanical properties.

465

Specimen	$R_e$ (MPa)	$R_{ms}$ (MPa) (at 7 % strain)
Reference	257	351
WD	248	393
BD	261	397
GD	232	378

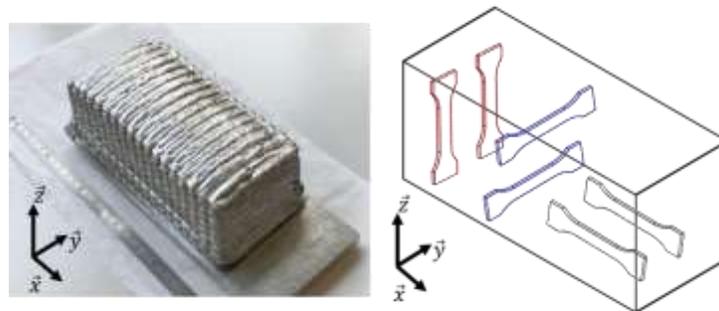
466

Table 4: mechanical properties for WD, GD and BD samples.

467

468 To complete this study, tensile tests, at room temperature, were carried out on flat specimens,  
469 whose geometry and dimensions were chosen according to ISO 6892-1 (gauge length of 33.54 mm,  
470 gauge width of 10 mm and nominal thickness of 2.5 mm). To achieve these tests, the specimens  
471 were extracted from a block sample (length: 250 mm, height: 80 mm width: 100 mm) manufactured  
472 with the CMT-P process. Two specimens for each extraction direction (x-direction or longitudinal  
473 direction, y-direction or transverse direction and z-direction or normal direction) were tested  
474 (Figure 7).

475

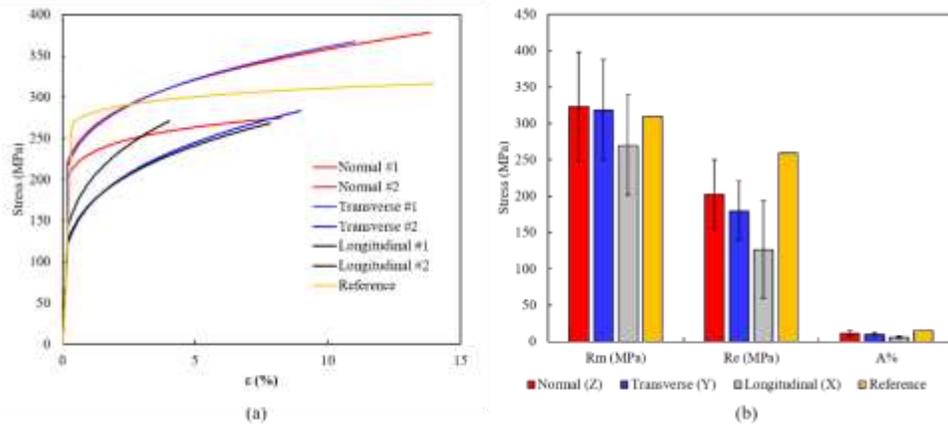


476

477

Figure 7: locations for extraction of samples for mechanical tests.

478  
 479  
 480 In Figure 8, we can see that one of the specimens collected in the longitudinal direction has a very  
 481 low elongation compared to the other specimens. A significant quantity of porosity was observed  
 482 on the failure surface after the mechanical test. However, we do not know the cause of this  
 483 increased porosity (proximity to the substrate, unmonitored change in electrical parameters...).



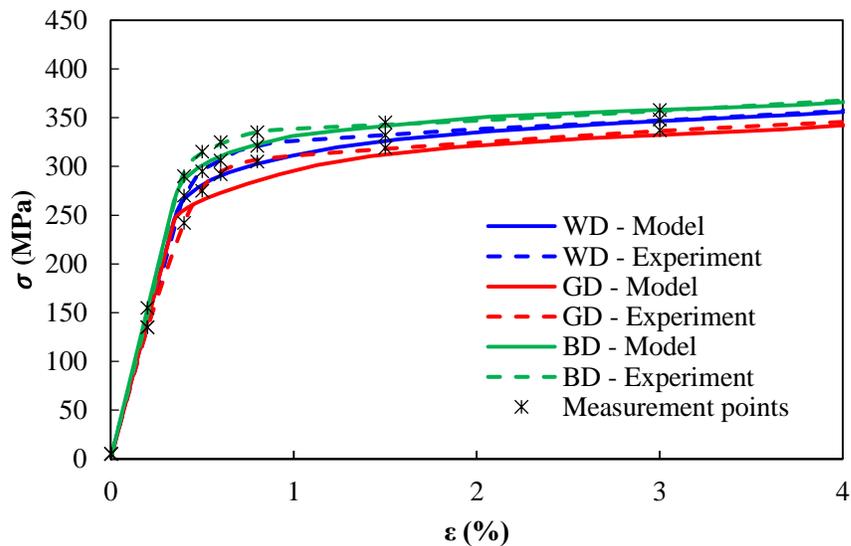
484  
 485 Figure 8: (a) stress–strain curves for tensile tests; (b): comparison of yield strength (Re), ultimate  
 486 strength (Rm) and elongation data of WAAM samples (reference sample extracted from a rolled  
 487 plate in T6 condition is also included for comparison).

488  
 489 The measured tensile properties are different due to a number of porosity higher than that of the  
 490 thin wall samples whatever the testing direction ( $\approx \pm 150$  MPa for Rm and  $\approx \pm 100$  MPa for Re).  
 491 Nevertheless, the tensile properties of the specimens in the normal (Z) and transverse (Y) directions  
 492 in the block sample are close to those measured in the walls (see Figure 8.b) and in the sample  
 493 obtained from a rolled 6061-T6 plate. These results confirm the possibility to obtain large-scale  
 494 aluminum parts with good mechanical properties. Additional investigations on the process  
 495 parameters should be made to obtain block sample with lower porosities like AA6061 thin-wall  
 496 samples.

497  
498 At the macroscopic scale, simulations are in good agreement with the experimental mechanical  
499 behavior (Figure 9). They give results close to the experimental values with a maximum stress  
500 deviation lower than 2 %. The elastic regime is, moreover, quite well reproduced. The choice of  
501 the single set of elastic constants from pure aluminum can therefore be validated at this scale of  
502 observation. At the macroscopic scale, the three experimental tensile curves display differences  
503 between them. At the onset of plasticity, the curve obtained with the WD specimen is about 5 %  
504 above the curve from the GD sample. Similarly, the curve from BD specimen is about 13 % higher  
505 than the curve from GD sample. The model closely reproduces this difference: the simulated WD  
506 curve is also about 5% above the predicted GD curve and the simulated BD curve is about 12 %.  
507 However, simulations slightly underestimate the mechanical behavior during the elastoplastic  
508 transition.

509

510



511 Figure 9: measured (dashed lines) and calculated (solid lines) macroscopic stress vs. macroscopic  
512 strain for the three specimens.

### 513 III.4. Comparison between *in situ* diffraction and EPSC modeling

514 The model gives access to the strain/stress state of each grain in the diffracting volume for a given  
515 macroscopic strain. It is possible to compare the experimental mechanical behavior at the  
516 diffracting volume scale, measured during *in situ* X-ray diffraction tensile experiments, to the  
517 simulated microscopic mechanical behavior. Varying the angles  $\varphi$  and  $\psi$  enables to probe different  
518 grain groups that have experienced a different deformation history depending on their orientation  
519 with respect to the tensile direction.

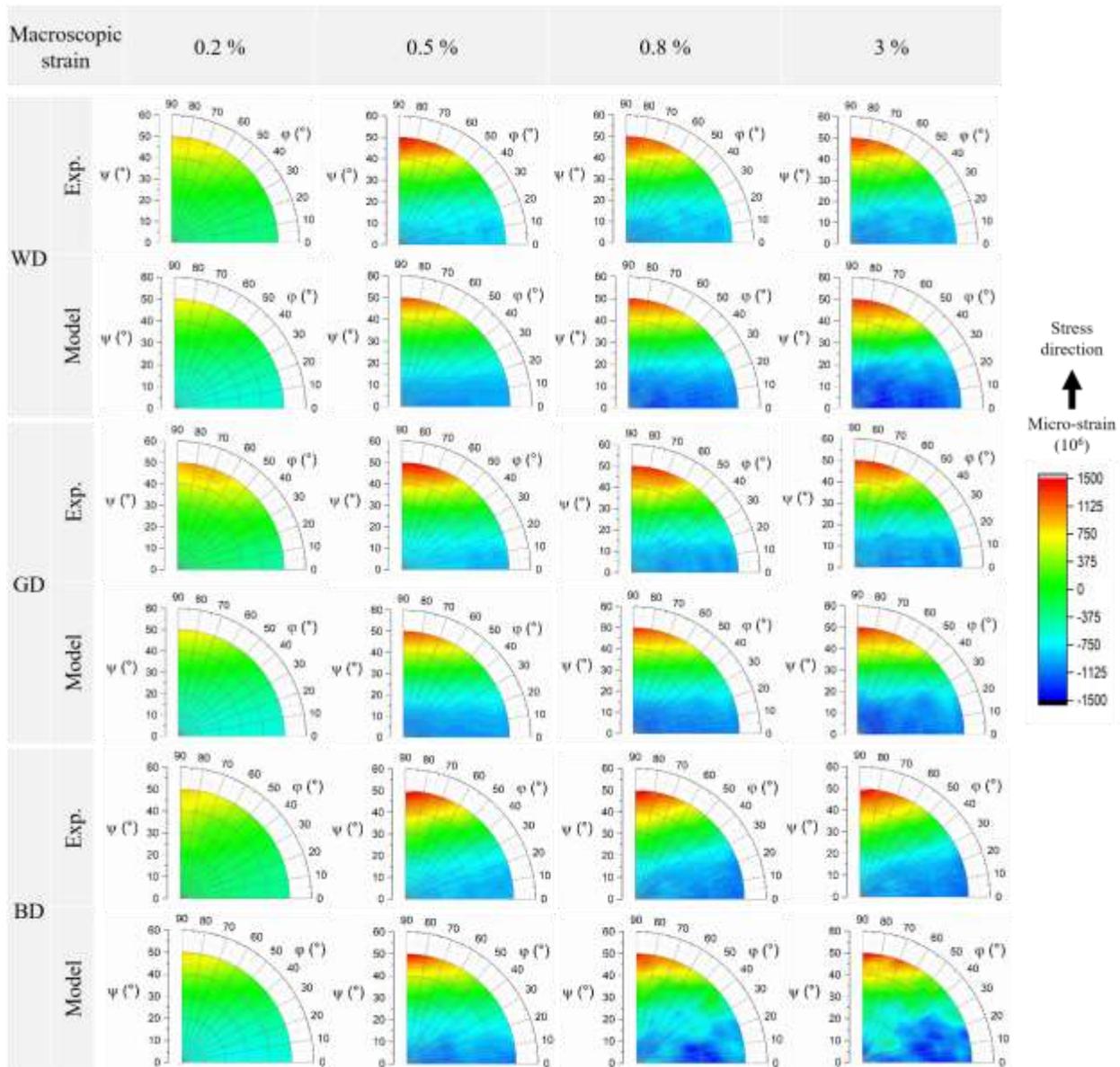
520 Predicted and experimental lattice strain pole figures for the {331} plane families are shown in Fig.  
521 10 for 4 macroscopic strains (0.2, 0.5, 0.8 and 3 %). Let us first focus on the development of the *in*  
522 *situ* measured elastic lattice strains for the BD sample and on a particular orientation, for example  
523 ( $\varphi = 90^\circ$ ;  $\psi = 50^\circ$ ). The strain pole figures calculated with the EPSC model display the same  
524 features as the experimental results. The elastic domain is very well reproduced. For example, the  
525 measured (respectively predicted) lattice strain at 0.2 % macroscopic strain is 802  $\mu\epsilon$  (resp.  
526 796  $\mu\epsilon$ ). During the elastoplastic transition (0.5 % macroscopic strain), the simulated values  
527 deviate slightly from the experimental ones. The predicted strain is about 10 % larger than the  
528 measured value. Finally, at 3 % macroscopic strain, the difference reaches 6 %.

529 The discrepancies between measurements and model calculations are more pronounced  
530 perpendicularly to the tensile axis ( $\varphi = 0^\circ$ ;  $\psi = 0^\circ$ ). In the elastic domain, the predicted and  
531 experimental lattice strain are, respectively -520  $\mu\epsilon$  and -300  $\mu\epsilon$ . These differences decrease with  
532 increasing strain. During the elastoplastic transition, the simulation shows a deviation of about  
533 20 % (i.e., about -1040  $\mu\epsilon$ ) from the experiment (-855  $\mu\epsilon$ ). At 3 % strain, the deviations are around  
534 16 %. Different authors have already highlighted that these discrepancies between simulations and  
535 measurements are more pronounced perpendicularly to the loading axis [35–37].

536 For the WD and GD specimens, the predicted results reproduce well the behavior identified  
537 experimentally using XRD (Figure 10). In the tensile direction ( $\varphi = 90^\circ$ ;  $\psi = 50^\circ$ ), the simulated  
538 strains are very close to those determined with XRD but they remain systematically lower than 7  
539 % whatever the applied strain for both WD and GD specimens. The elastic domain (0.2 %  
540 macroscopic strain) is also well reproduced: the measured (respectively predicted) lattice strains  
541 are: 847  $\mu\epsilon$  (resp. 803  $\mu\epsilon$ ) for the WD specimen. It can be noted that the smallest discrepancies  
542 between model and experiment are about 3 % at 3 % macroscopic strain. The maximum deviation  
543 observed in the tensile direction ( $\varphi = 90^\circ$ ) is 7 % and is reached in the elastoplastic transition for  
544 both specimens (e.g. 1280  $\mu\epsilon$  for the simulated lattice strains against 1372  $\mu\epsilon$  for the measured  
545 lattice strain for the GD specimen).

546 In the direction perpendicular to the tensile axis (i.e.  $\varphi = 0^\circ$ ;  $\psi = 0^\circ$ ), higher discrepancies are  
547 observed. They do not exceed 16 % in any case. The largest discrepancies (14 % for the GD  
548 specimen and 16 % for the WD specimen) are observed during the elastoplastic transition. In  
549 elasticity, the deviations are smaller (less than 10 %) for both specimens. In the plastic domain, at  
550 3 % macroscopic strain, the differences obtained are very weak (e.g. -1191  $\mu\epsilon$  for the simulated  
551 lattice strains against -1146  $\mu\epsilon$  for the measured lattice strains for the WD specimen). The model  
552 correctly describes the macroscopic mechanical behavior as well as the elastic strain field in the  
553 diffracting volume whatever the considered specimen. All the predicted values in elasticity are  
554 relatively close to the experimental ones, validating the choice made for the pure aluminum single-  
555 crystal elastic constants for the simulations.

556



557  
558  
559 Figure 10: experimental and simulated lattice strain pole figures for WD, BD and GD specimens.  
560

561 To explain the discrepancies between experimental results and simulation, some explanations can  
562 be proposed. More significant deviations of the simulated results from the experimental data can  
563 be observed in the direction perpendicular to the tensile axis To explain this point, it has to be noted  
564 that the grains which contribute to a given  $\{hk.l\}$  reflection in the loading direction involve only  
565 similarly oriented grains with respect to this direction and are thus not so sensitive to slight

566 variations in the representation of the crystallographic texture, contrary to  $\{hk.l\}$  grain groups in  
567 the direction perpendicular to the tensile axis. All grains with a plane-normal vector colinear to the  
568 scattering vector of the transverse direction, for a given  $\{hk.l\}$  reflection, have a shared  $(hk.l)$  plane-  
569 normal perpendicular to the loading axis. The grains which contribute to the reflection display  
570 many different crystallographic orientations with respect to the loading axis. In other words, a  
571 rotation of a grain around an axis perpendicular to the loading direction highly changes the stiffness  
572 in the tensile direction and, consequently, the strain and stress state in the grain. Different authors  
573 have already highlighted that these discrepancies between simulations and measurements are more  
574 pronounced perpendicularly to the loading axis [35–37]. The microscopic scale has not been  
575 explicitly described in the model (only one scale transition and the contribution of particles to  
576 single-crystal straining and hardening is not considered). In this study, the local behaviour (at the  
577 grain level) is based on the assumption of uniform mechanical fields. This approach permits to  
578 predict correctly the yield surface, the crystallographic texture linked to the plastic deformation,  
579 the intergranular stresses due to the initial and induced anisotropy during monotonic loading. To  
580 describe with more accuracy the metallurgical state at the grain level, a micro-mechanical  
581 description of a single crystal containing particles needs to be developed through a micro–meso  
582 transition. A modified Schmid's law and a new hardening matrix, taking into account the usual  
583 dislocation–dislocations interactions and also interactions between dislocations and particles, can  
584 be proposed. Next, a meso–macro transition using the self-consistent method proposed in the  
585 present work can be applied to deduce the global response of the polycrystalline material.

586

## 587 **Conclusions**

588

589 The mechanical properties of parts built by the WAAM process have been studied at the different  
590 scales of the material. The findings from the current study can be summarized as follows:

- 591 • Tensile tests carried out on samples extracted from the WAAM walls along three different  
592 orientations did not reveal relevant significant differences and thus demonstrate the quasi-  
593 isotropic character of the mechanical properties of the built material. The presence of  
594 equiaxed grains has probably reduced the anisotropic effect of columnar grains. Mechanical  
595 properties of the deposited material are slightly better as compared to conventional rolled  
596 material in T6 state.
- 597 • The residual stresses were characterized at the surface and in the bulk using diffraction.  
598 Although the residual stress values are low on the close-surface (40 MPa for the maximum  
599 value), the mechanical state is different in depth with significant maximum residual stress  
600 levels (+/- 100 MPa).
- 601 • A coupled method using a mean field approach and *in situ* mechanical XRD tests has been  
602 developed in order to better understand the development and the evolution of internal lattice  
603 strains under mechanical loading, considering the main microstructural features. The EPSC  
604 model enables to reproduce accurately the XRD strain pole figures as a function of the  
605 applied macroscopic deformation for the three sampling directions.

606

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