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1 Experimental and numerical investigation of ductile
2 fracture for AA6061-T6 sheets at room and elevated
3 temperatures

4 Ahmed Kacem^{a,*}, Hervé Laurent^a, Sandrine Thuillier^a

5 ^a*Univ. Bretagne Sud,*
6 *UMR CNRS 6027, IRDL, F-56100 Lorient, France*

7 **Abstract**

8 Warm forming is largely employed to enhance the poor formability of alu-
9 minum alloys at room temperature. The knowledge of the effect of stress state
10 and temperature on the forming limits is an important aspect for the control
11 of warm forming operations. Therefore, the objective of this work is to in-
12 vestigate the influence of temperature on the ductile fracture of AA6061-T6
13 aluminum alloy sheet metal under different stress states, ranging from shear
14 to biaxial tension. For this purpose, experiments and numerical simulations
15 of uniaxial tensile tests on dog-bone shaped specimen, notched specimens
16 with different radius, specimens with a central hole and shear specimens are
17 conducted at room and elevated temperatures. The hybrid experimental-
18 numerical approach is used to identify the fracture strain and the corre-
19 sponding stress state parameters (i.e. stress triaxiality and Lode parameter)
20 to derive the fracture loci at room temperature, 150 °C and 200 °C. To ac-
21 curately model the material behavior, the positive strain rate sensitivity in
22 the flow stress response at elevated temperature is considered and attention
23 is paid to constrain the finite element model by using the real experimental

*Corresponding author

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1 boundary conditions. A strong dependency of the ductile fracture on the
2 temperature and stress state is evidenced. Also, an extension of Lou's duc-
3 tile fracture criterion which includes the impact of temperature on ductile
4 fracture prediction is proposed. The prediction accuracy of the original and
5 extended criteria is evaluated. It is found that Lou's ductile fracture criterion
6 is well suited to predict accurately ductile fracture at elevated temperatures
7 under isothermal conditions whereas the extended form is very useful to pre-
8 dict ductile fracture initiation under non-isothermal conditions.

9

10 *Keywords:* Temperature, AA6061-T6, Ductile fracture, Stress state

11 **1. Introduction**

12 Aluminum alloys are extensively used in the metal forming industry due
13 to their high strength to weight ratio and excellent corrosion resistance. How-
14 ever, their poor formability at room temperature limits their use and makes
15 them less attractive than steel. Warm forming is increasingly used to enhance
16 the failure limits and therefore to manufacture complex structural compo-
17 nents made from aluminum alloys. Moreover, the large variety of stress states
18 encountered during a forming process must be taken into account. Indeed, it
19 is now well known that the forming limits depend strongly on the stress state
20 which may be described by the stress triaxiality η and the Lode parameter
21 L defined by:

$$\eta = \frac{\sigma_m}{\bar{\sigma}} \quad \text{with} \quad \sigma_m = \frac{\sigma_1 + \sigma_2 + \sigma_3}{3} \quad (1)$$

$$L = \frac{2\sigma_2 - \sigma_1 - \sigma_3}{\sigma_1 - \sigma_3} \quad (2)$$

1 where σ_m is the hydrostatic stress, $\bar{\sigma}$ is von Mises equivalent stress and σ_1 ,
 2 σ_2 and σ_3 are the principal stresses of the stress tensor, such that $\sigma_1 \geq \sigma_2 \geq \sigma_3$.

3 Hancock and Mackenzie [1] found out that the formability depends markedly
 4 on the stress triaxiality via a series of tests carried out on circumferentially
 5 notched tensile specimens. Mirza et al. [2] investigated the effect of stress
 6 triaxiality on the fracture characteristics of ductile metals for high stress tri-
 7 axiality level. It was evidenced that the ductility of all tested materials is
 8 strongly dependent on the stress triaxiality, this dependency being higher
 9 for the ferrous materials than for aluminium alloys. Bao and Wierzbicki [3]
 10 investigated the dependence of ductile fracture on the stress triaxiality in a
 11 wide range of this parameter by a series of tests including upsetting tests,
 12 shear tests and tensile tests on AA2024-T351 aluminum alloy, showing the
 13 strong dependency between formability and stress triaxiality. Zhang et al. [4]
 14 claimed that besides stress triaxiality, the Lode parameter should be consid-
 15 ered to give a complete description of the stress state. Wierzbicki et al. [5] and
 16 Gao and Kim [6] observed that predictions of ductile failure can be improved
 17 by introducing the Lode parameter.

18 Forming limits are not only related to the stress state, but also depend
 19 on temperature and strain rate. Shehata et al. [7] investigated the forma-
 20 bility of aluminum-magnesium alloys at temperatures ranging from 20 °C to
 21 300 °C, for several strain rates in both uniaxial and biaxial tension, by mea-
 22 suring the total elongation in tensile testing and the maximum cup height
 23 in punch stretching tests. They found out that, in the biaxial tension defor-

1 mation mode, the material is less sensitive to temperature and strain rate
2 than in uniaxial tension. Ayres and Wenner [8] investigated the forming
3 limits of AA5182-O by performing tensile and cup tests up to failure at
4 25 °C, 130 °C and 200 °C at different strain rates. It was concluded that
5 the limit strains increase at elevated temperature leading to higher form-
6 ing limit diagrams. Clausen et al [9] showed a considerable increase in the
7 fracture strain of AA5083-H116 at 400 °C, compared to room temperature.
8 Li and Ghosh [10] performed biaxial warm forming test for three automo-
9 tive aluminum sheet alloys, namely AA5757, AA5182 and AA6111-T4, in
10 the temperature range 200-350 °C. It was found out that all tested alloys
11 exhibit a significant improvement of their formability and the magnitude of
12 this effect depends on the aluminum grade. Mahabunphachai and Koç [11]
13 evidenced through tensile and bulge tests that the formability of AA5052
14 and AA6061 sheets increases with temperature and decreases with strain
15 rate. Similar observations are made by Chu et al. [12] for AA5086 using a
16 Marciniak test setup. Additionally, Wang et al. [13] showed that temperature
17 and punch speed have a significant effect on the formability through a cup
18 punch test. They also showed that the forming limits of AA2024 could be
19 improved considerably by forming in the vicinity, but not in excess, of 450 °C.
20 Recently, Kacem et al. [14] showed a strong dependency of the ductile fracture
21 of AA6061 sheets on the temperature and stress state.

22 Ductile Fracture (DF) criteria are widely used to assess the metal forma-
23 bility limits. The prediction of ductile fracture can be achieved by either an
24 uncoupled fracture criterion or a coupled damage model. Simplicity and a
25 reduced number of model parameters to be calibrated are the main advan-

1 tages of the uncoupled ductile fracture criteria, making them very attractive
 2 for industrial applications. In such an approach, the onset of fracture is pre-
 3 dicted when the integral of plastic deformation, multiplied by a weighting
 4 function depending on the stress state, reaches a critical value. By using dif-
 5 ferent weighting functions, several DF criteria based on this approach have
 6 been proposed in the literature at room temperature as shown in Tab.1.

DF criterion	Formulation
Rice and Tracey	$D(\bar{\varepsilon}_p) = \frac{1}{C_R} \int_0^{\bar{\varepsilon}_f} \exp\left(\frac{3}{2}\eta\right) d\bar{\varepsilon}_p$
Ayada	$D(\bar{\varepsilon}_p) = \frac{1}{C_A} \int_0^{\bar{\varepsilon}_f} \eta d\bar{\varepsilon}_p$
Lou	$D(\bar{\varepsilon}_p) = \frac{1}{C_3} \int_0^{\bar{\varepsilon}_f} \left(\frac{2}{\sqrt{L^2+3}}\right)^{C_1} \left(\frac{(1+3\eta)}{2}\right)^{C_2} d\bar{\varepsilon}_p$

Table 1: Some typical examples of DF criteria (Rice and Tracey [15], Ayada [16] and Lou [17, 18]). D is a failure indicator that gives the onset of fracture when it reaches unity. $\bar{\varepsilon}_p$ is the equivalent plastic strain. $\bar{\varepsilon}_f$ is the fracture strain. C_R , C_A , C_1 , C_2 and C_3 are material parameters.

7 Rice and Tracey [15], Oyane [19], Ayada et al. [16] and Ko et al. [20] pro-
 8 posed DF criteria that consider only the stress triaxiality effect on the ductile
 9 fracture. Bai and Wierzbicki [21, 22] proposed the Modified Mohr-Coulomb
 10 (MMC) criterion which considers the effect of both the stress triaxiality and
 11 Lode parameter. Mohr and Marcadet [23] developed the Hosford-Coulomb
 12 model considering also the effects of the stress triaxiality and the Lode param-
 13 eter. Lou et al. [17] developed a ductile fracture criterion inspired from the
 14 micro-mechanisms of ductile fracture, that occurs mainly due to void nucle-
 15 ation, growth and finally coalescence into microcracks. It was assumed that

1 void nucleation is proportional to the equivalent plastic strain, void growth is
2 controlled by the stress triaxiality and coalescence of voids is dominated by
3 the shear-linking up of voids governed by the maximum shear stress linked
4 to the Lode parameter.

5 The ductile fracture depends also on the strain rate and temperature.
6 However, the above-mentioned DF criteria do not consider this dependency
7 making them only suited for isothermal and fixed strain rate conditions.
8 Some ductile fracture criteria are extended to include the effect of tempera-
9 ture and/or strain rate, which makes it possible to account for fracture under
10 non-isothermal and changing strain rate conditions. Novella et al. [24] pro-
11 posed an extension of Oyane's criterion with the introduction of a function of
12 temperature and strain rate, obtained by a linear interpolation on the basis
13 of the critical damage values obtained in tension for six conditions of temper-
14 ature and strain rate. This criterion was calibrated by means of tensile tests
15 conducted on smooth specimens only and applied to a cross wedge rolling
16 process carried out on AA6082-T6 bars at elevated temperature. Roth and
17 Mohr [25] extended the Hosford-Coulomb DF criterion by adding the strain
18 rate dependency based on an analogy with the Johnson-Cook criterion [26] to
19 investigate ductile fracture initiation in advanced high strength steel sheets.
20 In a similar way, Liu et al. [27] extended the Bao-Wierzbicki fracture model [3]
21 by adding the strain rate and the temperature dependency to simulate the
22 chip separation in a cutting process of AA2024-T351 cylindrical parts. It
23 was found that the extended model is the most suitable to describe the chip
24 removal behavior of ductile materials. Also in the same manner, Du et al. [28]
25 extended the MMC criterion to investigate the fracture behavior of 3.2 mm

1 thick AA5383 sheet at high temperatures, spanning from 350 to 450 °C. Wang
2 et al [29] proposed a new DF criterion by adding a temperature-dependent
3 parameter in the MMC criterion, to describe the fracture behavior of AZ31
4 magnesium alloy sheets as a function of the temperature and stress state.
5 Fracture tests were carried out at room and elevated temperatures using
6 different specimen geometries. However, the fracture strain was measured
7 experimentally on the specimen surface by Digital Image Correlation (DIC)
8 technique, that is not enough to accurately calibrate the fracture criterion
9 since the fracture may initiate within the thickness rather than on the surface
10 in this kind of experiments. More recently, Cao et al. [30] extended Lou's DF
11 criterion [31] by including temperature and strain rate terms in a similar way
12 to Johnson-Cook criterion to study the fracture behavior of AA7075-H112
13 bulk metal.

14 Figuring out the relationship between forming limits, temperature, strain
15 rate and stress state is of great importance for providing more understanding
16 of ductile fracture in warm forming conditions. However, from the above-
17 mentioned studies, it is evident that there is still a need for a further inves-
18 tigation of the effect of stress state and temperature on the forming limits
19 of aluminum alloy sheets. The main aim of this work is to study experi-
20 mentally and numerically the ductile fracture of an aluminum alloy sheet for
21 a wide range of stress states, i.e. from shear to biaxial tensionn at room
22 and elevated temperatures. To this regard Lou's DF criterion is selected to
23 describe the fracture locus and evaluate the forming limits of AA6061-T6 alu-
24 minum alloy at room temperature (RT), 150 °C and 200 °C. Indeed, Lou and
25 Huh [32] compared and evaluated the fracture loci constrained by various DF

1 criteria at room temperature. The comparison demonstrated that the MMC
2 and Lou DF criteria provide the best predictability in a wide range of stress
3 states. Moreover, Lou’s DF criterion has been successfully applied for pre-
4 dicting fracture of several aluminum alloys and therefore seems well adapted
5 to predict the onset of fracture in metal forming processes (e.g. AA1050 [33],
6 AA6082-T6 [34] and AA6016-T4 [35]). The ductile fracture criterion is cali-
7 brated by the hybrid experimental-numerical approach [36, 37]. To this end,
8 experiments and numerical simulations of uniaxial tensile tests on dog-bone
9 shaped specimen, notched specimens with different radius, specimens with a
10 central hole and shear specimens are conducted at room and elevated temper-
11 atures. Since an accurate description of the flow behavior is fundamental in
12 the hybrid approach, the positive strain rate sensitivity at elevated tempera-
13 ture and the initial anisotropy of the sheet are considered. Attention is also
14 paid to accurately constrain the finite element model by using the real ex-
15 perimental boundary conditions measured by DIC. The aim is to determine
16 the fracture strain and the corresponding stress state parameters for each
17 specimen geometry corresponding to a given stress state and temperature.
18 Furthermore, an extension of Lou’s DF criterion is suggested to include the
19 effect of temperature on the fracture initiation. The ductile fracture locus
20 obtained by both the original and extended Lou’s DF are compared to the
21 experimental results.

22 **2. Material and methods**

23 This section firstly details the material and specimens designed to obtain
24 the different stress states. Then, details of experiments and FE modeling of

1 ductile fracture tests are described. Finally, DF criterion and the methodol-
2 ogy used to calibrate its parameters are presented.

3 *2.1. Material and specimens*

4 AA6061-T6 aluminum alloy, which is widely used for structural compo-
5 nents in the automotive industry, is selected for this study.

6 Five different types of specimens illustrated in Fig.1 are cut from the same
7 1 mm aluminum sheet using Computer Numerical Control (CNC) waterjet
8 cutting machine. Dog-bone specimen are designed according to ISO 6892-1
9 standard to conduct monotonic tensile test. Four ductile fracture specimens
10 are chosen in such a way to achieve a wide range of stress states. Notched
11 specimens with small (NR5) and large (NR15) notch radii are considered to
12 achieve a high stress triaxiality [38]. Specimen with a central hole (HR4)
13 is designed to obtain a stress state close to the uniaxial tensile test but
14 with minor necking [36]. Finally, shear specimen (SH) is designed to achieve
15 a low stress triaxiality [39]. Moreover, dog-bone specimens are cut in the
16 rolling direction (RD), as well as in 45° and 90° to RD, to investigate the
17 anisotropy. Ductile fracture specimens are only extracted along RD. Three
18 samples are prepared for each fracture specimen type and direction.

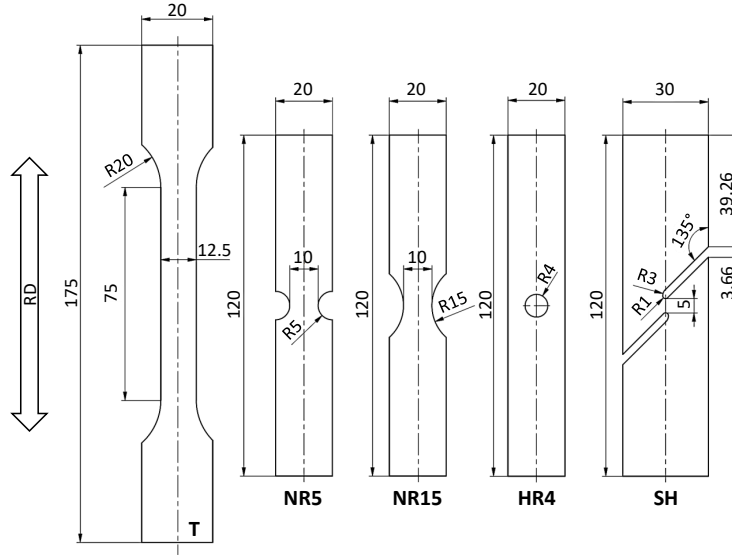


Figure 1: Geometry and dimensions (in mm) of specimens designed to obtain different stress states, (T) dog-bone specimen designed for monotonic tensile test according to ISO 6892-1 standard, (NR5) notched specimen with $R = 5$ mm, (NR15) notched specimen with $R = 15$ mm, (HR4) specimen with a central hole with $R = 4$ mm and (SH) shear specimen. All specimens are extracted from 1 mm AA6061-T6 sheet along RD. Dog-bone specimens are also extracted along 45° and 90° to RD.

1 2.2. Tensile and ductile fracture tests

2 All tests are performed with an INSTRON 5969 universal testing machine
 3 equipped with a 50 kN load cell under isothermal conditions ($T = RT$, 150°C
 4 and 200°C).

5 A classical furnace (Matair) is used to heat up the specimen to the test-
 6 ing temperatures of 150°C and 200°C . The temperature is measured by two
 7 K-type thermocouples attached at the center of the specimen and the grip
 8 of the tensile machine respectively. To insure a homogenous temperature on

1 the specimen during the test, the tensile load is applied when the tempera-
2 ture recorded by both thermocouples reaches the testing temperature within
3 $\pm 2^\circ\text{C}$. To avoid any thermal stresses due to thermal expansion before ap-
4 plying the tensile load, the load is maintained close to zero by moving the
5 crosshead manually during the heating stage. Two tests among three are
6 selected for each specimen geometry for the post-processing.

7 Tensile tests are conducted under extensometer-based strain control at a
8 constant strain rate of $\dot{\epsilon} = 10^{-3} \text{ s}^{-1}$. To this end, a temperature-resistant
9 mechanical extensometer with a 25 mm gauge length is used. Fracture tests
10 are also conducted in such a way to obtain the same strain rate employed
11 for the tensile test. However, strain rate control cannot be used in ductile
12 fracture test since the deformation is not homogeneous in the gauge length
13 area of the extensometer. Indeed, the local strain rate at the fracture initi-
14 ation site may be 10 times higher than in the rest of specimen. Therefore,
15 the tests are carried out under displacement control at a constant crosshead
16 velocity, which is set to obtain an average strain rate of $\dot{\epsilon} = 10^{-3} \text{ s}^{-1}$.

17 The deformation evolution during experiments is recorded using Digital
18 Image Correlation. To prepare specimens for DIC, the sample surface is
19 cleaned and spray-painted with a white and black stochastic pattern using a
20 spray can with acrylic paint. Two high speed cameras equipped with 50 mm
21 lenses are used to acquire the images for the DIC. Despite the attention paid
22 to ensure the alignment of the specimen with respect to the loading direction,
23 a slight misalignment of the sample during the test is likely to occur in this
24 kind of experiments, as was reported by Kacem et al.[40] and Tang et al.[41].
25 Indeed it is impossible to ensure a complete alignment during the test, and

1 even a slight misalignment will cause great rotation and asymmetry in the
2 process of large plastic deformation leading to an asymmetrical strain distri-
3 bution as was observed experimentally in many previous works (e.g. Lou et
4 al. [34], Deole et al. [42] and Gruben et al. [43]). To account for this misalign-
5 ment, five virtual extensometers with a gauge length of 15 mm are defined
6 at different positions of the gauge area of specimen, using the DIC software
7 Aramis [44] as shown in Fig.2 for the testing of a specimen with a central
8 hole. Fig.2 shows also the evolution of the displacements of the five virtual
9 extensometers in function of the normalized time. It can be observed that
10 the displacement strongly depends on the position of the virtual extensome-
11 ter. Therefore, for all specimen geometries, the values are averaged and the
12 local displacement u corresponds then to the average of the displacements
13 recorded by the five virtual extensometers.

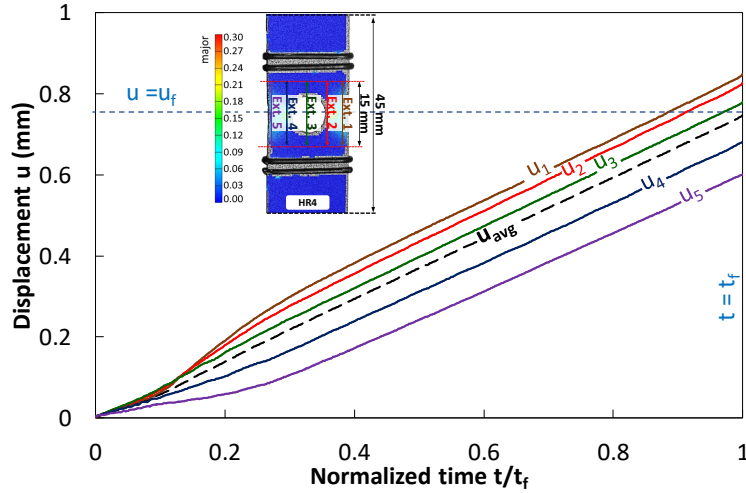


Figure 2: Evolution of displacement of the five virtual extensometers, with a gauge length of 15 mm, defined at different positions with DIC software as a function of time t normalized by the fracture time t_f , for a specimen with central hole HR4. The dashed black line depicts the average value of the five virtual extensometer displacements and u_f corresponds to the displacement at fracture.

1 2.3. Numerical models

2 Numerical simulations of the fracture tests are performed in 3D with the
3 finite element code Abaqus/Standard. Since the strain is localized in the
4 specimen center and DIC measures give the local displacement field, only a
5 reduced part with a length $L = 45$ mm is modelled, as shown in Fig.3. In
6 addition, to develop a FE model as close as possible to the experimental
7 conditions, a model without symmetry conditions is defined to account for
8 all possible displacements of the specimen during the test.

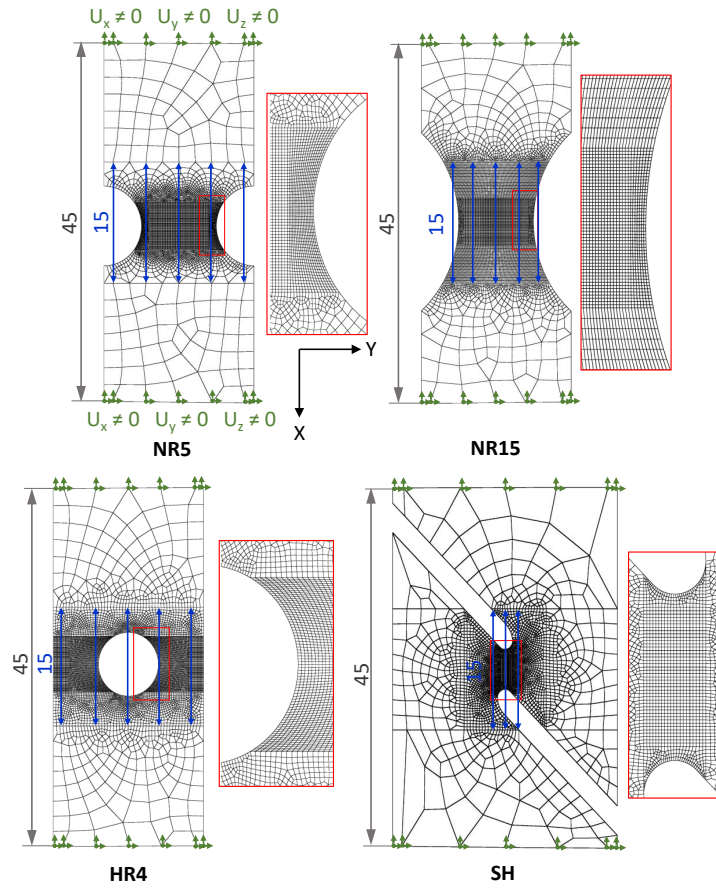


Figure 3: Meshes in X-Y plane and boundary conditions used for the four ductile fracture tests. Eight elements are defined through the thickness. The blue arrows highlight the position of the virtual extensometers with a gauge length of 15 mm used for local displacement measurement. All dimensions are in mm.

- 1 The region of interest is meshed using 3D solid hexahedral elements with
- 2 reduced integration (C3D8R). A fine mesh is used in the local zone where
- 3 strain localization is likely to occur and the size is increased at sections away
- 4 from this zone. A mesh size of approximately 0.1 mm is applied to the fine

1 mesh area on the basis of a mesh sensitivity study showing that the effect of
 2 mesh size becomes insignificant when it is smaller than this value. Moreover,
 3 8 layers of hexahedral elements are used through the thickness.

4 Recently, Kacem et al. [40] have shown that experimental boundary con-
 5 ditions have a significant effect on the fracture strain evaluation. Therefore,
 6 real non-zero displacement components (U_x , U_y and U_z) are imposed on the
 7 lower and upper sides of the specimen, with values derived from the experi-
 8 mental displacement field recorded by the DIC software at the same area, as
 9 shown in Fig.3. To this end, an automatic procedure is developed to apply
 10 the experimental local displacements to the FE model with a Python script
 11 as detailed in Appendix A. Such a procedure leads to a more realistic model,
 12 in which the misalignment of the specimen during the test is reproduced. For
 13 all simulations, isothermal conditions are assumed.

14

15 Isotropic hardening coupled with Hill48 yield criterion is chosen to model
 16 the mechanical behavior. The Hill48 quadratic yield function is given by:

$$f = \bar{\sigma} - \sigma_y(\bar{\varepsilon}_p) \quad (3)$$

17 where $\bar{\sigma}$ is the Hill48 equivalent stress that is defined as follow, from the
 18 components of the Cauchy stress tensor:

$$\bar{\sigma} = \sqrt{F(\sigma_{yy} - \sigma_{zz})^2 + G(\sigma_{zz} - \sigma_{xx})^2 + H(\sigma_{xx} - \sigma_{yy})^2 + 2L\sigma_{yz}^2 + 2M\sigma_{xz}^2 + 2N\sigma_{xy}^2} \quad (4)$$

19 The Hill's coefficients (F , G , H , L , M and N) are related to the plastic
 20 anisotropy ratios, or r-values, (r_0 , r_{45} and r_{90}) which are the ratios of the

1 width to thickness incremental plastic strain during a tensile test at 0° , 45°
 2 and 90° to RD, respectively:

$$G = \frac{1}{1 + r_0} \quad H = Gr_0 \quad F = \frac{Gr_0}{r_{90}} \quad N = G \left(\frac{r_0}{r_{90}} + 1 \right) (r_{45} + 0.5) \quad (5)$$

3 Due to the lack of available data regarding the mechanical behavior in
 4 the thickness of the sheet, L and M are kept equal to their isotropic value
 5 (i.e $L = M = 1.5$).

6 At elevated temperatures, the viscosity becomes significant. The strain
 7 rate dependency is commonly introduced as a power law in the isotropic
 8 hardening function σ_y , that is chosen with a saturation form of Hockett-
 9 Sherby, e.g. [45, 46]. Thus, the yield stress σ_y is defined at each temperature
 10 studied as a function of the strain rate $\dot{\varepsilon}$ as:

$$\sigma_y(\bar{\varepsilon}_p, \dot{\varepsilon}) = [\sigma_0 + Q (1 - \exp(-b (\bar{\varepsilon}_p)^n))] \left[\frac{\dot{\varepsilon}}{\dot{\varepsilon}_0} \right]^m \quad (6)$$

11 where $\bar{\varepsilon}_p$ is the equivalent plastic strain, σ_0 is the initial yield stress, Q
 12 represents the maximum change in the size of the yield surface, b defines
 13 the growth rate of the yield surface, n is the strain hardening coefficient,
 14 $\dot{\varepsilon}_0$ is a constant strain rate normalization factor and m is the strain rate
 15 sensitivity coefficient. It should be noted that for all strain rate below the
 16 strain rate normalization factor $\dot{\varepsilon}_0$, the yield stress σ_y is calculated only from
 17 the first bracketed term of Eq. 6. To define the hardening law in Abaqus, the
 18 stress-strain data are given as input in a tabular form at different strain rate
 19 levels.

1 *2.4. Modeling and calibration of ductile fracture*

2 *2.4.1. Ductile fracture criterion*

3 A stress state dependent ductile fracture criterion is used to determine the
4 onset of the macroscopic fracture. This criterion characterizes the damage
5 state in the material and when a critical value is reached, it leads to the onset
6 of a macroscopical fracture. This criterion can be expressed according to the
7 following general form:

$$D = \int_0^{\bar{\varepsilon}_f} \frac{d\bar{\varepsilon}_p}{\bar{\varepsilon}_f} \quad (7)$$

8 where D is a failure indicator that gives the onset of fracture at a given
9 material point when it reaches unity and $\bar{\varepsilon}_f$ is the equivalent plastic strain
10 to fracture (or fracture strain) that can be expressed according to Lou et
11 al. [17, 18] as follows:

$$\bar{\varepsilon}_f = \frac{C_3}{\left(\frac{2}{\sqrt{L^2+3}}\right)^{C_1} \left(\frac{(1+3\eta)}{2}\right)^{C_2}} \quad (8)$$

12 C_1 , C_2 and C_3 are three positive material parameters that need to be
13 calibrated.

14 *2.4.2. Calibration methodology*

15 To calibrate the fracture material parameters, the equivalent plastic strain
16 at the first element reaching $D=1$ at the instant of fracture t_f needs to
17 be determined. However, this element cannot be located since the fracture
18 parameters required to calculate D have not been identified yet. Therefore,
19 it is assumed that the fracture initiates in the element with the highest
20 equivalent plastic strain at the instant of fracture t_f (critical element). This

1 assumption had been checked and broadly applied in previous works. Indeed,
2 Talebi-Ghadikolaee et al. [47] found that, in the case of notched specimens,
3 the position of the element with the highest equivalent plastic strain is the
4 same as that of the element with the highest failure indicator value. Lou and
5 Huh [48] have shown that the element with the maximum failure indicator is
6 close to the one with the largest equivalent plastic strain for specimen with
7 central hole and shear specimen.

8 Moreover, in most experiments, fracture is not expected to initiate on
9 the specimen surface making the direct measure of fracture strain through
10 DIC impossible. Therefore, the hybrid experimental–numerical method is
11 applied [36]. To this end, the displacement at fracture u_f is first deter-
12 mined experimentally at the instant of the sudden drop in the measured
13 load-displacement curve. Subsequently, finite element simulations of ductile
14 fracture tests are performed and the evolution of the equivalent plastic strain,
15 stress triaxiality and Lode parameter are output. After verifying the agree-
16 ment between experimental and numerical results, the maximum equivalent
17 plastic strain is extracted from the numerical simulation of ductile fracture
18 test at the corresponding experimental fracture stroke u_f . The evolution of
19 stress triaxiality η and Lode parameter L at the critical element is also ob-
20 tained from the numerical analysis, to account for the non-linearity in the
21 loading path, when identifying the fracture parameters. Therefore, the fail-
22 ure indicator D at the critical element can be calculated for each test through
23 Eq. 7.

24 Finally, to calibrate the fracture criterion, the fracture parameters are
25 optimized using a generalized reduced gradient (GRG) algorithm tool by

1 minimizing the following cost function defined in the least square sense:

$$Erf = \sum_{i=1}^{N_s} (1 - D_i)^2 \quad (9)$$

2 where N_s denotes the number of samples used for the calibration and
3 D_i corresponds to the failure indicator of the i^{th} test which is calculated
4 numerically through Eq. 7.

5 **3. Results and discussion**

6 In this section, results of tensile and ductile fracture tests are presented
7 and discussed. The effect of temperature on the stress state and fracture
8 strain is investigated. The results of DF parameters calibration of the original
9 Lou's DF at different temperatures are presented. Then the extended form
10 of Lou's DF criterion is described. The ductile fracture locus obtained by
11 the original and extended forms are compared to the experimental results.

12 *3.1. Stress-strain curves*

13 Cauchy stress-logarithmic strain curves at room and elevated tempera-
14 tures ($T = 150^\circ\text{C}$ and 200°C) are obtained from tests performed on dog-bone
15 specimen as shown in Fig.4. A representative curve is chosen among the three
16 ones and it should be highlighted that very repeatable data are obtained for
17 each temperature. The Cauchy stress is calculated by the ratio of the load
18 over the current section of the sample before necking. The logarithmic strain
19 is measured by the DIC software. The mechanical properties are given in
20 Tab. 2. Constant values for elasticity parameters ($E = 70 \text{ GPa}$ and $\nu = 0.33$)
21 are assumed for all tested temperatures.

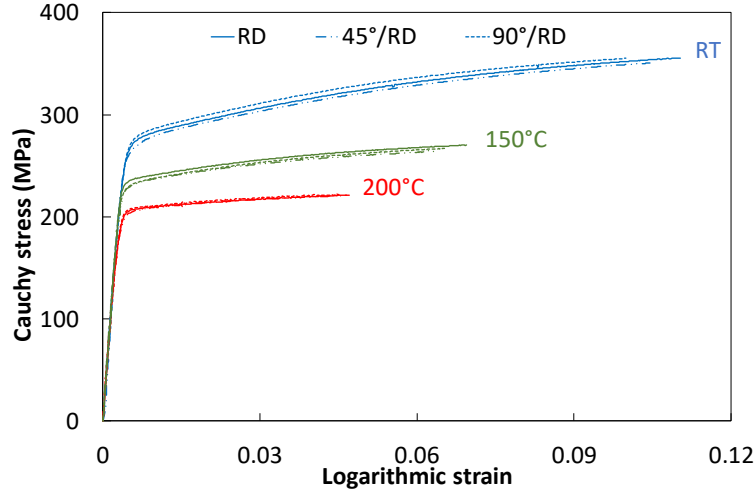


Figure 4: Cauchy stress-logarithmic strain curves of AA6061-T6 at $\dot{\epsilon} = 10^{-3} \text{ s}^{-1}$ in uniaxial tension obtained with dog-bone specimen up to the point of necking at 0° , 45° and 90° to the RD, at room temperature and at 150°C and 200°C .

	YS (MPa)	UTS (MPa)	r_0	r_{45}	r_{90}	\bar{r}	Δr
RT	270	318	0.59	0.78	0.81	0.74	-0.08
150°C	235	252	0.59	0.66	0.72	0.66	-0.005
200°C	205	211	0.64	0.71	0.72	0.7	-0.03

$$\bar{r} = (r_0 + r_{90} + 2r_{45})/4 \quad \Delta r = (r_0 + r_{90} - 2r_{45})/2$$

Table 2: Mechanical properties in RD and anisotropy coefficients of AA6061-T6 at room temperature and at 150°C and 200°C .

- 1 It can be seen that, for all tested temperatures, there is no significant
- 2 difference in stress-strain curves along different loading directions indicating
- 3 isotropic properties with respect to strength and strain hardening, though

1 a slight gap recorded at RT that clearly disappears at higher temperatures.
2 However, the flow behavior is significantly influenced by the temperature. As
3 excepted, the Ultimate Tensile Stress (UTS) and the Yield Stress at 0.2%
4 plastic strain (YS) decrease with the temperature rise. For instance, UTS
5 and YS decrease respectively by 33.6% and 24.1% when increasing the tem-
6 perature from RT to 200 °C. It can be observed that the limit strain of the
7 homogeneous deformation field decreases with increasing the temperature.
8 This is due to the work hardening that decreases as the temperature in-
9 creases leading to a faster occurrence of necking.

10 The measurement of both transverse and longitudinal strains leads to the
11 calculation of the plastic anisotropy coefficients at 0°, 45° and 90°, also given
12 in Tab. 2. These coefficients are determined by linear interpolation of the evo-
13 lution of the plastic transverse strain versus the plastic thickness strain in
14 the whole range of homogeneous deformation. The transverse strain is mea-
15 sured by DIC and the thickness strain is determined using the incompress-
16 ibility condition; moreover, plastic components are calculated by removing
17 the elastic part. Fig.5 shows the r-values evolution with the orientation from
18 rolling direction at different testing temperatures. It is observed that for all
19 temperatures, the value of the normal anisotropy coefficient is significantly
20 different from 1, which indicates that the metal exhibits moderate anisotropy
21 in plastic flow.

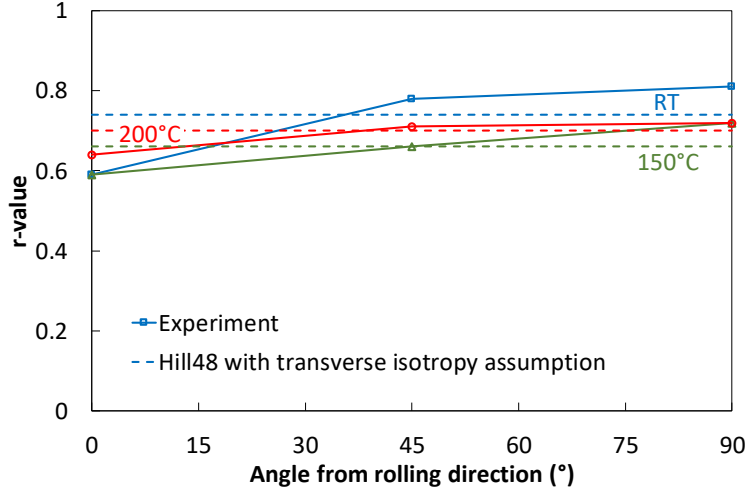


Figure 5: Plastic anisotropy coefficients (r-values), determined experimentally by the ratio of the plastic transverse strain to the plastic thickness strain in the whole range of homogeneous deformation, in function of orientation and temperature and comparison with the transverse isotropic Hill48 yield criterion calculated according to Eq.5 and considering only \bar{r} (i.e. $r_0 = r_{45} = r_{90} = \bar{r}$).

1 3.2. Identification of mechanical behavior parameters

2 The Hill48 yield criterion parameters are identified directly from the r-
 3 values. Due to the very weak dependence of the stress-strain curve on the
 4 loading direction and the low value of the planar anisotropy coefficient Δr ,
 5 transverse isotropy is assumed and the values of Hill48 coefficients, calculated
 6 according to Eq.5 and considering only \bar{r} (i.e. $r_0 = r_{45} = r_{90} = \bar{r}$), are reported
 7 in Tab.3.

8 The hardening parameters are identified by inverse method. To this end,
 9 two strategies are commonly used. The first strategy consists in manually ad-
 10 justing the unknown model parameters while visually checking the agreement

	F	G	H	N
RT	0.575	0.575	0.425	1.425
150 °C	0.602	0.602	0.398	1.398
200 °C	0.588	0.588	0.412	1.412

Table 3: Hill’s coefficients calculated directly through r-values according to Eq.5 and considering only \bar{r} (i.e. $r_0 = r_{45} = r_{90} = \bar{r}$) at room temperature and at 150 °C and 200 °C. $L = M = 1.5$

1 between experimental and numerical results. The second strategy consists
2 in using an automatic process aiming at seeking for an optimal set of model
3 parameters by minimizing an objective function value that defines the gap
4 between experimental and numerical results. For its simplicity, the first strat-
5 egy is used in this study. Indeed, the hardening parameters are continually
6 adjusted until the numerical and experimental load-displacement curves ob-
7 tained with all fracture specimens show good agreement as shown in Fig. 7.
8 Tab. 4 shows the set of hardening parameters, obtained after several trial-
9 and-error iterations and leading to the numerical results in Fig. 7. It should
10 be noted that the hardening parameters are commonly identified based on
11 the results of tensile test in the homogeneous strain range. After necking
12 a classical approach consists in extrapolating the hardening curve up to a
13 large strain range. Nevertheless, in this work, the hardening curve param-
14 eters are identified based on the load-displacement curves obtained with all
15 fracture specimens up to fracture while checking the agreement between the
16 experimental and numerical stress-strain curve obtained by tensile test in the

1 homogeneous strain range.

	σ_0 (MPa)	Q (MPa)	b	n	$\dot{\epsilon}_0$ (s^{-1})	m
RT	244	135	7	0.68	–	–
150 °C	164	100	7	0.45	10^{-4}	0.031
200 °C	130	76	7	0.36	10^{-4}	0.04

Table 4: Hardening law parameters identified at room temperature and at 150 °C and 200 °C.

2 It should be also noted that in practice the strain rate sensitivity param-
3 eters (i.e. $\dot{\epsilon}_0$ and m) are identified based on experimental results for different
4 strain rates. However in this study experiments are conducted at only one
5 strain rate value of $\dot{\epsilon} = 10^{-3} s^{-1}$. Therefore these parameters are constrained
6 to be as close as possible to the ones found in previous works for the same
7 material grade [46]. Moreover, the strain hardening exponent n is assumed to
8 evolve linearly with the temperature, as was reported previously by Laurent
9 et al. [45] and the parameter b is fixed to be constant for all tested temper-
10 atures. It is found that both hardening parameters σ_0 and Q decrease with
11 the temperature rise by about 45%. The σ_0 parameter evolves in a linear
12 manner with temperature. The strain rate parameter m increases with the
13 temperature rise, indicating an increase of the positive strain rate sensitivity
14 with temperature, as was also reported by Simoes et al. [49].

15 Fig. 6 shows the hardening curves obtained at $\dot{\epsilon} = 10^{-3} s^{-1}$ and using
16 the identified parameters as well as the hardening curve determined experi-
17 mentally from the tensile tests. It is clearly seen that the calibrated Hockett-

- 1 Sherby gives a good prediction of the stress level at $\dot{\epsilon} = 10^{-3} \text{ s}^{-1}$ within the
- 2 limited strain range investigated with the tensile test.

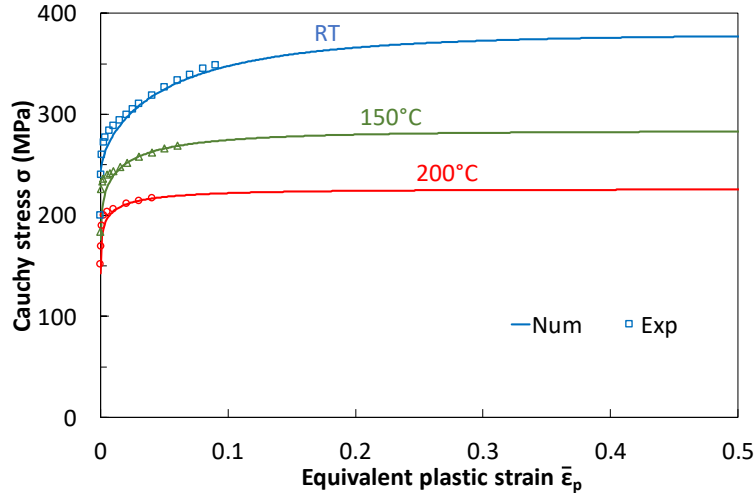


Figure 6: Hardening curves predicted by the modified Hockett-Sherby model at room temperature and at 150 °C and 200 °C (straight lines). Symbols represent experimental points obtained from uniaxial tension of dog-bone specimen at $\dot{\epsilon} = 10^{-3} \text{ s}^{-1}$ in RD.

- 3 A comparison of the experimental and predicted load-local displacement
- 4 curves is shown in Fig. 7 for the fracture tests. For clarity purpose, only one
- 5 test per configuration is plotted.

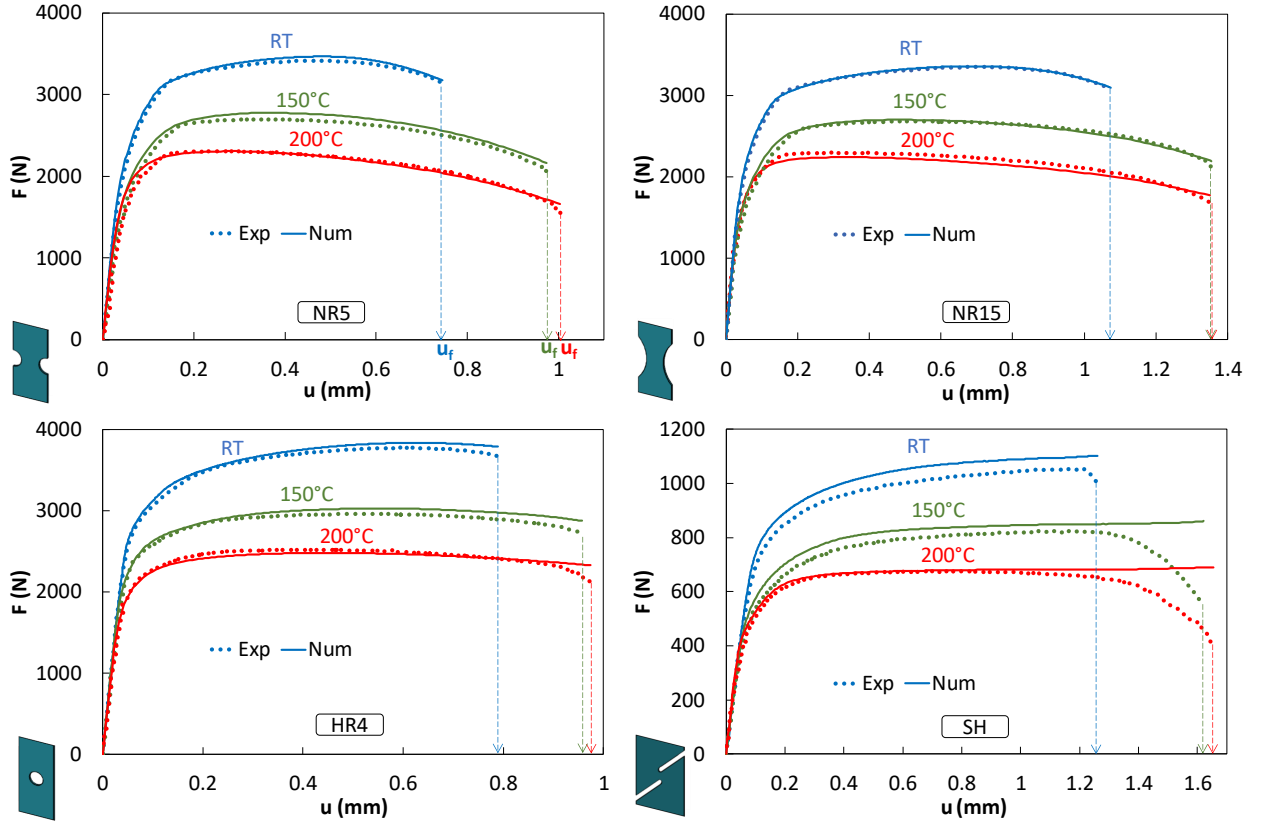


Figure 7: Numerical and experimental load-local displacement curves at room temperature and at 150 °C and 200 °C. Vertical dashed lines represent the displacement at fracture u_f . For clarity purpose, only one test per configuration is plotted.

1 Since this data is also used for the calibration, a good agreement is ob-
 2 served between numerical and experimental results. For shear test, a slight
 3 discrepancy of about 5% is observed between experimental and numerical
 4 results at room temperature that decreases with the raise of temperature.
 5 This can be attributed to the lack of flexibility of the quadratic yield cri-
 6 terion used in this work characterized by too few anisotropic parameters as

1 was noticed by Bron and Besson [50]. This can be improved by using another
2 yield function with more anisotropic parameters (e.g. Bron and Besson yield
3 function [50] and Yld2004-18p [51]) as was reported by Zhang et al. [52]. It
4 should be noted that in a previous work conducted for the same material at
5 room temperature [40], the load-local displacement curves in Fig. 7 are also
6 obtained numerically and experimentally for the same specimen geometries
7 but with neglecting the anisotropy and the strain rate dependency. However,
8 in this previous work, the shear specimen is used to identify the hardening
9 curve after necking. Therefore, a good agreement is obtained between exper-
10 imental and numerical results for the shear specimen but a slight difference
11 is observed in the case of notched specimen NR5.

12 It is found out that the temperature affects strongly the maximum load
13 F_{max} , namely it decreases with increasing temperature for all specimen ge-
14 ometries. At constant temperature, the force levels between both notched
15 specimens are quite similar while it is slightly larger for HR4 specimen and
16 drops significantly for the SH specimen. Indeed, at room temperature, the
17 maximum load is found to be respectively 3422 N and 3356 N for notched
18 specimens NR5 and NR15, 3791 N for HR4 specimen and 1069 N for SH
19 specimen. At elevated temperatures, the same trend is recorded. Also, it
20 can be seen that the maximum load is more rapidly reached at elevated tem-
21 peratures. For instance, for NR5 specimen, F_{max} is reached at 66% u_f and
22 23% u_f for RT and 200 °C, respectively. This is due to the localized necking
23 that appears earlier at elevated temperatures.

24 The numerical simulations are also verified by comparing the major strain
25 distribution just before fracture and the evolution of the average major strain

1 on the surface of the critical zone, where the fracture is most likely to occur for
2 each specimen as illustrated in Fig. 8. For elevated temperatures, experimen-
3 tal results are not plotted until fracture because the speckles spray-painted
4 on the surface of the critical zone of specimen peeled off due to the high
5 temperature and the large deformation prior to fracture, thus making impos-
6 sible the DIC computation. Only major strain distribution at RT are plotted
7 in Fig. 8 since DIC results are not available just before fracture. It can be
8 seen that the numerical results fit well the experimental ones. The asym-
9 metrical experimental strain distribution, coming from the misalignment of
10 the specimen during the test, is also well predicted by the numerical model
11 for all tested specimens and temperatures. The maximum major strain is
12 found close to the edge for NR5 specimen and close to the center for NR15
13 specimen. For HR4 specimen, the major strain is concentrated on the right
14 side of the hole edge. For SH specimen, the maximum major strain is located
15 close to the edge where fracture experimentally initiates.

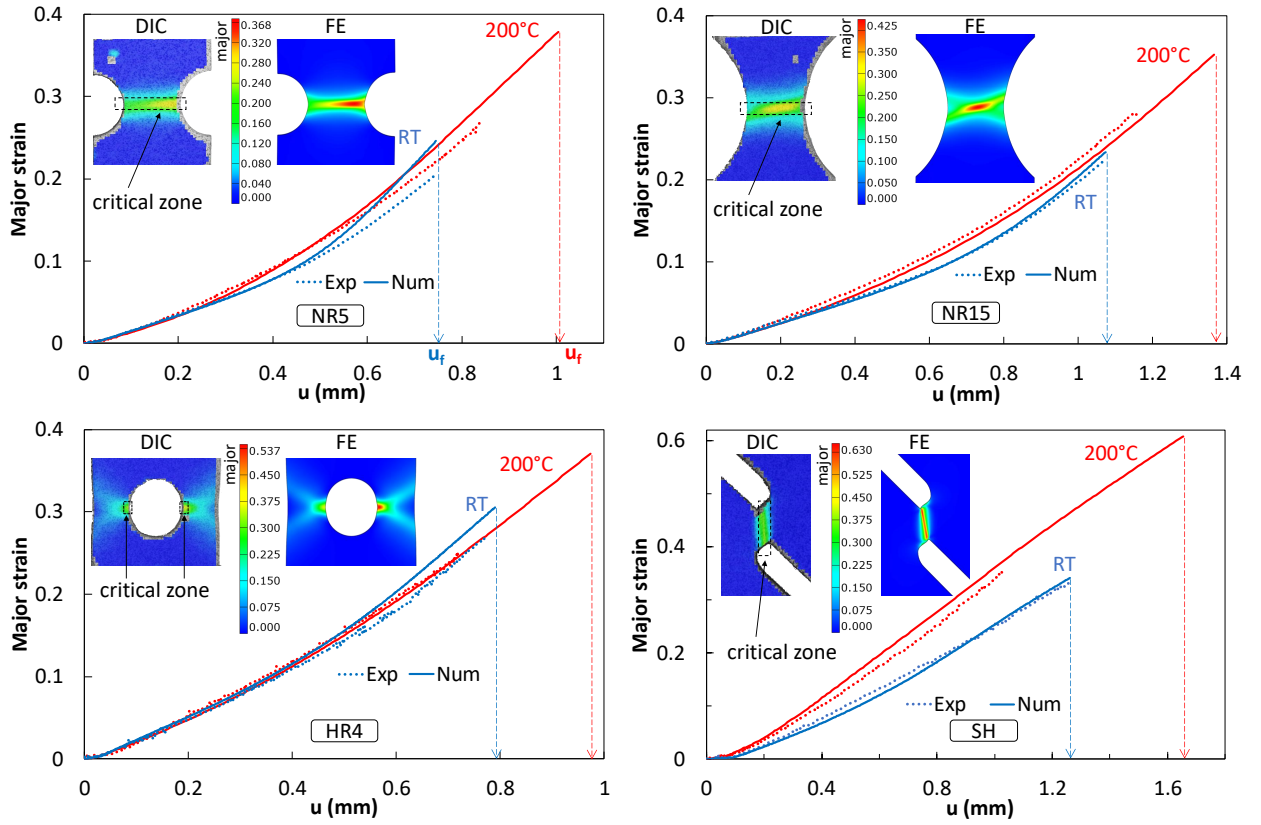


Figure 8: Comparison of major strain distribution on the specimen surface between DIC and FE simulation just before fracture at room temperature and the evolution of the average major strain on the surface of the critical zone where the fracture is likely to occur at room temperature and 200 °C. Vertical dashed lines represent the displacement at fracture u_f . Experimental results are not plotted until fracture at 200 °C because the speckles spray-painted on the surface of the critical zone of specimen peeled off due to the high temperature.

1 3.3. Effect of temperature on the stress state and fracture strain

2 Hybrid experimental-numerical approach is used to determine the fracture
 3 strain and the evolution of the stress state parameters (η and L) until the

1 fracture initiation. Fig. 7 shows the displacement at fracture u_f determined
 2 at the instant of the sudden drop in the measured load-displacement curve
 3 for NR5, NR15 and HR4 specimens. For SH specimen, the load-displacement
 4 curve gradually drops before fracture making the determination of u_f in this
 5 way more difficult. Therefore, a novel method that makes use of the load
 6 first derivative is adopted [29]. In this case, the fracture stroke is identified
 7 at the minimum of this curve as shown in Fig. 9.

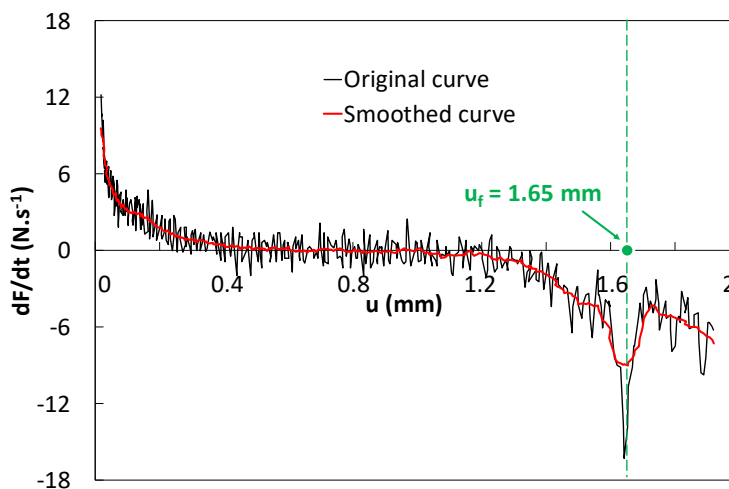


Figure 9: Original (in black) and smoothed (in red) load first derivative-**local displacement** curves of the shear specimen at 200 °C. Vertical dashed lines depict the minimum of the smoothed curve corresponding to the displacement at fracture u_f .

8 It can be seen that the displacement at fracture depends strongly on
 9 the specimen geometry. The maximum value of u_f is obtained with the SH
 10 specimen while the minimum value with NR5 and HR4 specimens. The tem-
 11 perature affects also the displacement at fracture u_f with less effect beyond
 12 150 °C. For instance, for NR5 specimen, u_f rises by 31% when increasing the

1 temperature from RT to 150 °C while it rises by 4.7% when increasing the
2 temperature from 150 °C to 200 °C. The value of u_f remains quite constant
3 when increasing the temperature from 150 °C to 200 °C for NR15 and HR4
4 specimens.

5 Fig. 10 shows the location of the element with the maximum equivalent
6 plastic strain at u_f . It is found that the critical element is located within
7 the thickness for NR5, NR15 and HR4 specimens. For SH specimen, due
8 to the edge effect, the maximum equivalent plastic strain at u_f is observed
9 at the edge for all tested temperatures. However, the stress state is not
10 characteristic of shear in this region. Therefore, the central element of the
11 shear gauge is chosen as the critical element rather than the element at the
12 edge. It is also found out that the location of the critical element is insensitive
13 to the temperature. Experimentally, when inspecting visually the specimen
14 after fracture, it was found that the fracture initiates close to the location of
15 maximum equivalent plastic strain.

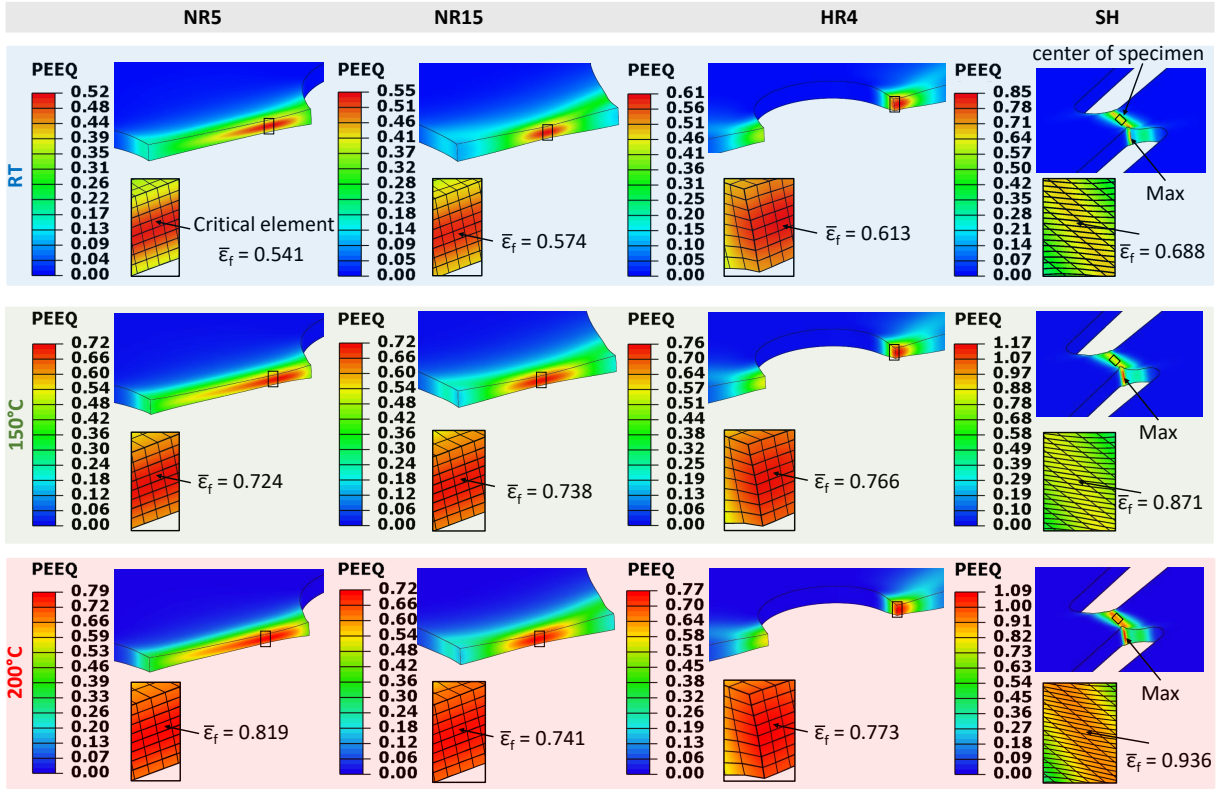


Figure 10: Contour plot of the equivalent plastic strain at the onset of fracture in the critical zone for different ductile fracture tests at room temperature and at 150 °C and 200 °C. Arrows in the zoomed pictures show the element with the maximum equivalent plastic strain corresponding to the critical element for NR5, NR15 and HR4 specimens and the element at the center of the specimen corresponding to the critical element for SH specimen. Note that the fracture strain $\bar{\epsilon}_f$ corresponds to the value of the equivalent plastic strain at the integration point of the critical element. Only one of the two post-processed tests per configuration is presented.

- 1 To visualize the loading paths to fracture, Fig. 11 shows the evolution of
- 2 stress triaxiality and Lode parameter calculated at the critical element as a
- 3 function of the equivalent plastic strain and temperature for each specimen

1 type.

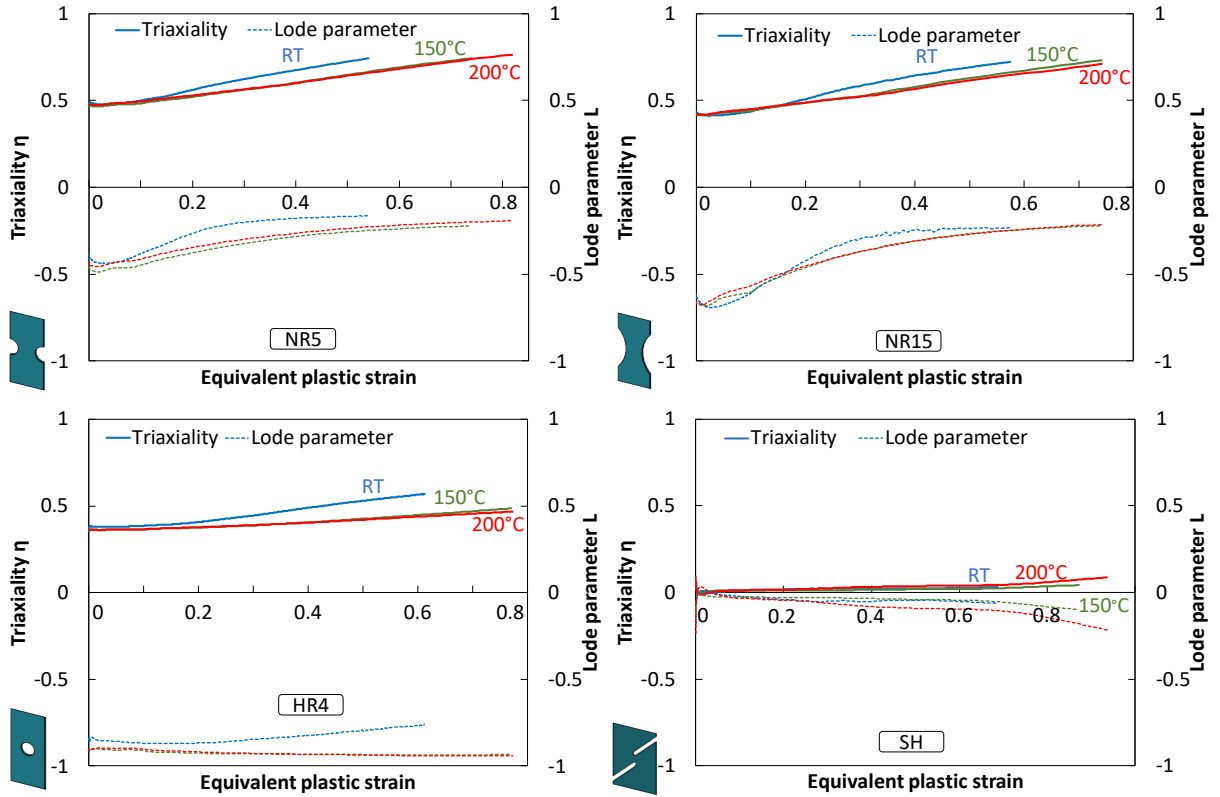


Figure 11: Evolution of stress triaxiality and Lode parameter at the critical element corresponding to the element presenting the maximum equivalent plastic strain at the displacement at fracture u_f for NR5, NR15 and HR4 specimens and selected at the center of the shear gauge for SH specimen. Only one of the two post-processed tests per configuration is plotted.

2 It is observed that the stress triaxiality values covered by the experi-
 3 ments are positive, ranging from 0 up to 0.75, while negative values of Lode
 4 parameters ranging from -1 up to 0 are obtained. Although specimens are
 5 designed to achieve a constant stress state, it is found that regardless of the

1 testing temperature the values of stress state parameters show significant
2 variation during plastic deformation, especially for the notched specimens.
3 This is due to the localized necking leading to the development of out of
4 plane stress components in the critical zone that increases the stress triaxi-
5 ality and modifies the evolution of Lode parameter. For NR5 specimens at
6 RT, the stress triaxiality rises from 0.48 at the start of plastic deformation
7 up to 0.71 at the fracture initiation. For NR15 specimens at RT, the stress
8 triaxiality increases from 0.42 to 0.72. Quite similar values at the start and
9 the end of deformation are also obtained at 150 °C and 200 °C. However, the
10 stress triaxiality rises more rapidly at RT since the plastic deformation at
11 fracture is lower than the one obtained at elevated temperatures. The same
12 trend is also observed for the Lode parameter, respectively rising from -0.38
13 to -0.17 and from -0.62 to -0.2 for NR5 and NR15 specimens at RT. For HR4
14 specimen, the variation of stress state parameters is smaller than the one
15 of notched specimens, since only a weak necking occurs. At RT, the stress
16 triaxiality and Lode parameter increase respectively from 0.36 to 0.52 and
17 from -0.86 to -0.82. However, nearly constant values are observed at elevated
18 temperatures. For SH specimen, a perfect pure shear state, corresponding
19 to $\eta = 0$ and $L = 0$, is not obtained but the values of η and L are nearly
20 constant and close to zero.

21 Since stress state parameters are evolving during deformation, it is con-
22 venient to assign a specific stress state to each fracture test by using the
23 average of stress triaxiality η_{avg} and Lode parameter L_{avg} calculated by the
24 following relations:

$$\eta_{avg} = \frac{1}{\bar{\varepsilon}_f} \int_0^{\bar{\varepsilon}_f} \eta d\bar{\varepsilon}_p \quad \text{and} \quad L_{avg} = \frac{1}{\bar{\varepsilon}_f} \int_0^{\bar{\varepsilon}_f} L d\bar{\varepsilon}_p \quad (10)$$

1 Tab.5 shows the values of the fracture strain $\bar{\varepsilon}_f$ and the corresponding
2 average triaxiality η_{avg} and Lode parameter L_{avg} .

		$\bar{\varepsilon}_f$	η_{avg}	L_{avg}
NR5	RT	0.469 ± 0.072	0.584 ± 0.019	-0.27 ± 0.011
	150 °C	0.697 ± 0.028	0.585 ± 0.01	-0.339 ± 0.019
	200 °C	0.807 ± 0.013	0.609 ± 0.003	-0.287 ± 0.002
NR15	RT	0.56 ± 0.015	0.563 ± 0.003	-0.382 ± 0.004
	150 °C	0.703 ± 0.036	0.558 ± 0.008	-0.382 ± 0.006
	200 °C	0.739 ± 0.003	0.562 ± 0.002	-0.366 ± 0.002
HR4	RT	0.576 ± 0.038	0.428 ± 0.029	-0.869 ± 0.035
	150 °C	0.739 ± 0.028	0.408 ± 0.003	-0.93 ± 0.001
	200 °C	0.74 ± 0.034	0.376 ± 0.031	-0.931 ± 0.004
SH	RT	0.655 ± 0.034	0.021 ± 0.002	-0.047 ± 0.005
	150 °C	0.857 ± 0.014	0.017 ± 0.002	-0.039 ± 0.005
	200 °C	0.901 ± 0.036	0.038 ± 0.003	-0.093 ± 0.004

Table 5: Fracture strain and the corresponding average stress triaxiality and Lode parameter with the absolute uncertainty at room temperature and at 150 °C and 200 °C. The absolute uncertainty is calculated based on the two post-processed tests for each geometry and temperature.

3 It can be seen that the fracture strain is strongly dependent on the stress
4 state and temperature. The fracture strain may increase by more than a

1 factor of 1.7 when increasing the temperature from RT to 200 °C (for NR5
 2 specimen) or by more than a factor of 1.4 when decreasing the stress triaxi-
 3 ality value from 0.58 to 0.02 (at RT).

4 Fig. 12 shows the evolution of η_{avg} and L_{avg} as a function of temperature.
 5 It can be seen that the temperature has less effect in case of the notched
 6 specimen NR15 where η_{avg} and L_{avg} remain at constant values respectively
 7 of about 0.56 and -0.38 regardless of the testing temperature. However, the
 8 stress state of the other specimens depends slightly on the testing temper-
 9 ature. For instance, for HR4 specimen, η_{avg} and L_{avg} decreases by about
 10 0.05 when increasing the temperature from RT to 200 °C. Therefore, It can
 11 be stated that the temperature effect on the stress state is rather small and
 12 depends on the specimen geometry.

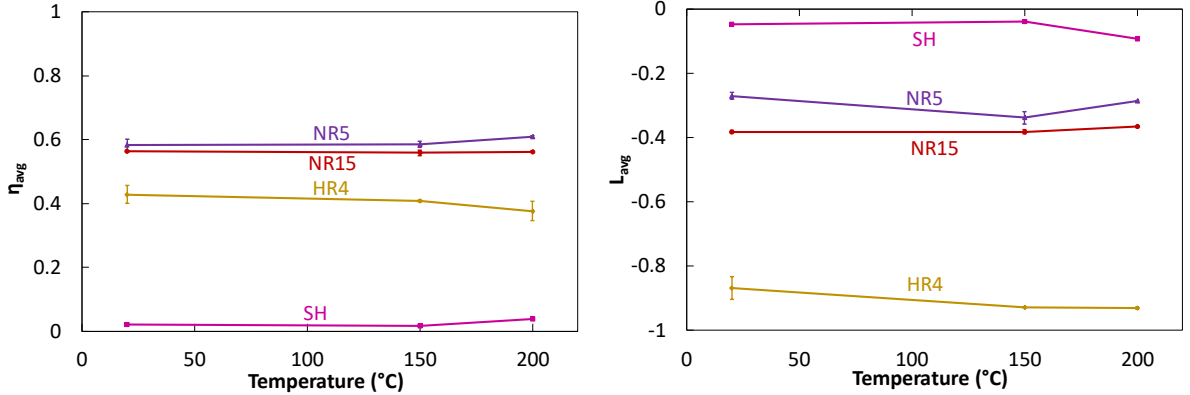


Figure 12: Evolution of the average triaxiality η_{avg} and Lode parameter L_{avg} , calculated according to Eq.10, as a function of temperature for all fracture specimens. Error bars depict the maximum and minimum values of the two post-processed tests for each geometry and temperature.

1 3.4. Prediction of ductile fracture

2 The ductile fracture criterion used in this study depends on three param-
3 eters (C_1 , C_2 and C_3) that have to be identified using the procedure described
4 in 2.4.2 to derive the fracture locus. Calibration is performed by using all
5 the post-treated tests for each temperature (i.e. $N_s = 4 \times 2 = 8$). Tab.6 shows
6 the calibration results. Fig.13 shows the evolution of the calibrated fracture
7 parameters as a function of temperature.

	C_1	C_2	C_3
RT	0.6398	0.2500	0.6118
150 °C	0.0877	0.2130	0.7562
200 °C	0	0.1662	0.8009

Table 6: Calibrated fracture parameters of Lou's ductile fracture criterion at RT, 150 °C and 200 °C.

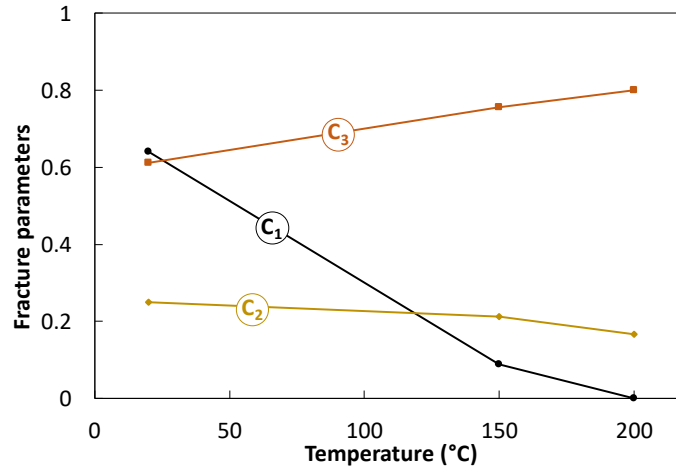


Figure 13: Evolution of the three calibrated Lou's fracture parameters (C_1 , C_2 and C_3) as a function of temperature.

1 It can be seen that the value of C_1 , which modulates the effect of the
2 Lode parameter, is small compared to the values found in literature (e.g.
3 $C_1 = 3.3845$ for AA6082-T6 [34] and $C_1 = 2.8703$ for AA6016-T4 [35] at RT)
4 for all tested temperatures. This indicates a weak Lode parameter depen-
5 dence for this material, notably at elevated temperatures, that was also noted
6 in previous studies for AA5083-H116 [53] and AA5083-O [54] aluminum al-
7 loys at RT. The values of the fracture parameter C_2 , which modulates the
8 effect of the stress triaxiality, indicates a moderate stress triaxiality sensitiv-
9 ity that decreases with increasing temperature in agreement with the results
10 of Pantousa et al. [55] for steel. As expected, the parameter C_3 , which in-
11 dicates the overall level of fracture strain and corresponds theoretically to
12 the fracture strain in uniaxial tension, increases with the temperature rise.
13 Indeed, C_3 increases by 31% when increasing the temperature from RT to
14 200°C. It should be noted that the calibrated values obtained at RT in this
15 study are slightly different from those found out in a previous work for the
16 same material grade ($C_1 = 0.1588$, $C_2 = 0.5725$ and $C_3 = 0.4593$) [40], since
17 the anisotropy is considered and more tests are used for calibration.

18 Fig.14 shows the fracture loci resulting from the calibration of the DF
19 criterion corresponding to the plot of fracture strain as the function of the
20 stress triaxiality and the Lode parameter. The symbols represent the fracture
21 strain obtained by the hybrid approach with their corresponding average
22 stress triaxiality and Lode parameter.

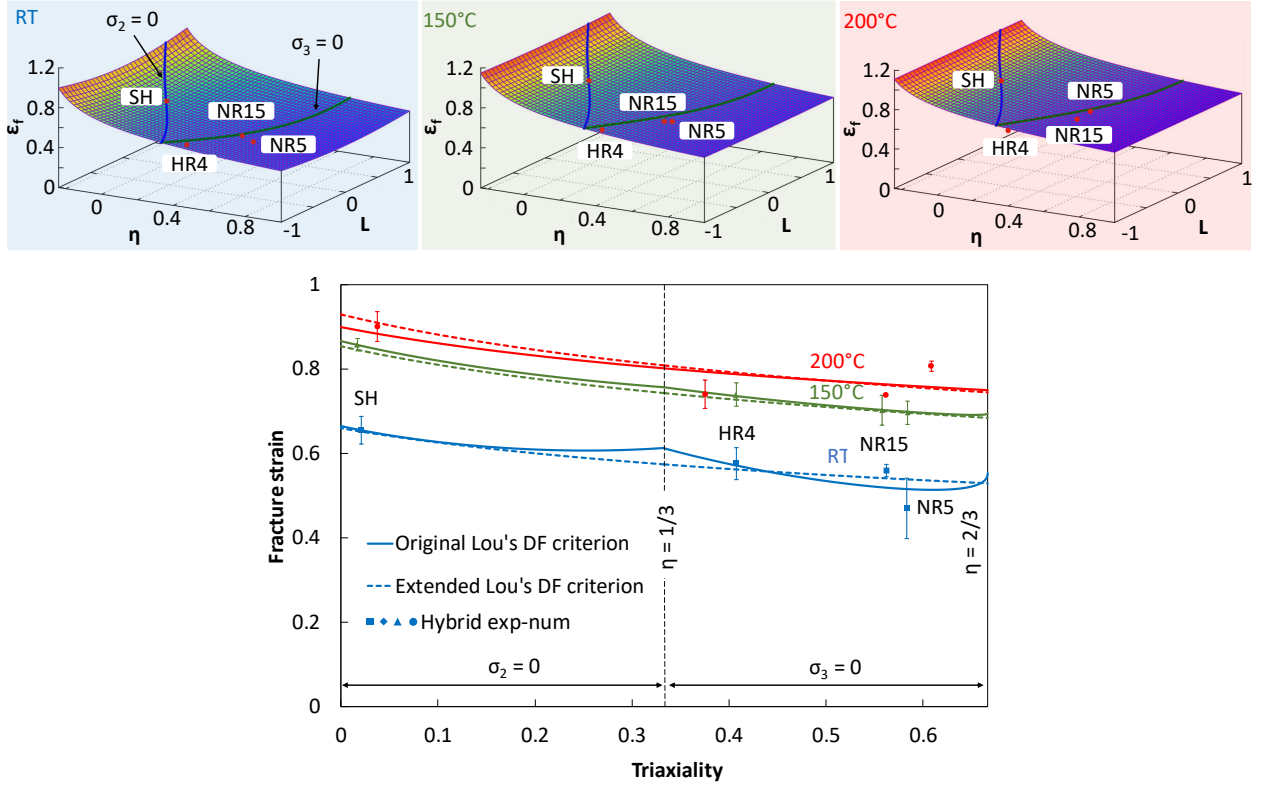


Figure 14: Fracture locus of AA6061-T6 constructed in the space of stress triaxiality, Lode parameter and equivalent plastic strain by the original Lou's DF criterion (top of the figure) and in the space of stress triaxiality and equivalent plastic strain under plane stress by the original and extended Lou's DF criterion (lower part of the figure) at RT, 150 °C and 200 °C. Dots with corresponding error bars represent $(\eta_{avg}, \bar{\epsilon}_f)$ points obtained by the hybrid experimental-numerical approach. Error bars depict the maximum and minimum values of fracture strain $\bar{\epsilon}_f$ of the two post-processed tests for each geometry and temperature.

- 1 The overall shape of the fracture loci obtained at 150 °C and 200 °C in
- 2 the top part of the figure is similar, where an evident stress triaxiality de-
- 3 pendence and weak Lode parameter dependence are observed. The shape of

1 the fracture locus obtained at RT is slightly different from the one obtained
 2 at elevated temperatures, where a much stronger dependence on the stress
 3 triaxiality and on the Lode parameter is seen. This is clearly seen in the
 4 lower part of the figure, where the fracture loci are plotted under the plane
 5 stress condition in the space of stress triaxiality and equivalent plastic strain.
 6 It should be noted that a much stronger dependence of the fracture surface
 7 on the stress state and temperature has been observed for other materials as
 8 was found, for instance, by Pandya et al. [56] for AA7075-W.

9 The $(\eta_{avg}, \bar{\epsilon}_f)$ points obtained by the hybrid experimental approach are
 10 superimposed in Fig.14. It can be seen that globally the locus fits all the data
 11 points obtained by the hybrid approach quite well. To evaluate more precisely
 12 the prediction accuracy of DF criterion parameters an error indicator is used:

$$\delta = \frac{|D_{cri} - D_{pred}|}{D_{cri}} \cdot 100\% \quad (11)$$

13 where D_{cri} is the critical value of D which is supposed to be one in this
 14 work and D_{pred} is the value of D in the critical element at the fracture stroke.

15 Tab.7 shows the prediction error for the ductile fracture criterion calibra-
 16 tion at room and elevated temperatures.

		NR5	NR15	HR4	SH
RT	D_{pred}	0.9027	1.0601	0.9929	1.0014
	error (%)	9.73	6.01	0.71	0.14
	mean error (%)	4.15			
150 °C	D_{pred}	0.9957	0.9979	0.9998	1.0005
	error (%)	0.44	0.21	0.02	0.05
	mean error (%)	0.18			
200 °C	D_{pred}	1.0655	0.9670	0.9332	1.0199
	error (%)	6.55	3.30	6.68	1.99
	mean error (%)	4.63			

Table 7: Failure indicator D predicted in the critical element at the fracture stroke by Lou’s DF criterion and prediction errors (Eq.11).

1 It is found that the mean errors of the three tested temperatures are small
2 (lower than 5%) notably at 150 °C, where the predicted results fit perfectly
3 the experimental ones. Therefore, it can be concluded that Lou’s ductile
4 fracture criterion widely used in previous works to predict DF initiation at
5 room temperature (e.g. [42, 57]) is also well suited to predict accurately DF
6 at elevated temperatures.

7 3.5. Extension of Lou’s ductile fracture criterion

8 Industrial warm forming operations are usually conducted under non-
9 isothermal conditions. However, the DF parameters are identified for a lim-
10 ited isothermal conditions. To overcome this limitation, in this section, an
11 extension of Lou’s DF criterion is proposed by adding a new parameter to
12 include the effect of temperature on the DF prediction.

1 Fig.15 shows the evolution of the predicted fracture strain by Lou's ductile
 2 fracture criterion with temperature for different stress states.

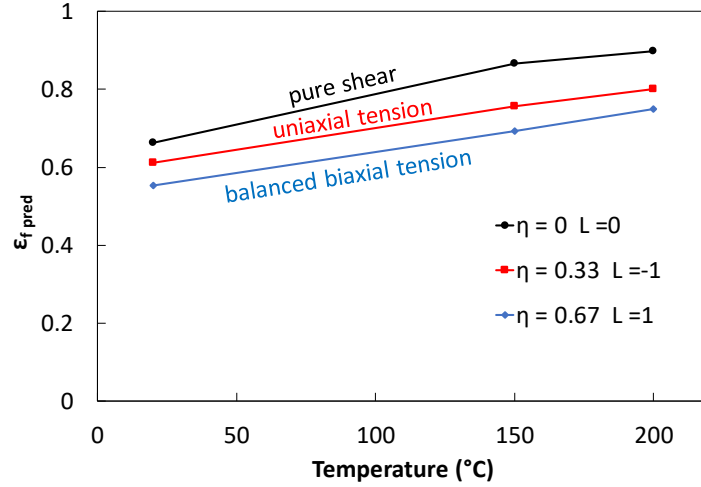


Figure 15: Evolution of fracture strain predicted by the original Lou's DF criterion with the calibrated material parameters given in Tab.6 versus temperature for three different stress states of pure shear ($\eta=0$, $L=0$), uniaxial tension ($\eta=0.33$, $L=-1$) and balanced biaxial tension ($\eta=0.67$, $L=1$).

3 It can be seen that the predicted fracture strain gradually changes in
 4 an approximately linear way for all stress states. The same trend was also
 5 noted in [26], and a ductile fracture model was developed, based on multi-
 6 plicative contributions considering the effects of stress state, strain rate and
 7 temperature:

$$\bar{\epsilon}_f = [C_5 + C_6 \exp(C_7 \eta)] [1 + C_8 \ln(\dot{\epsilon}/\dot{\epsilon}_0)] [1 + C_9 T^*] \quad (12)$$

8 where C_5 to C_9 denotes the Johnson-Cook fracture parameters and T^* is
 9 the homologous temperature, defined by :

$$T^* = \frac{T - T_{ref}}{T_m - T_{ref}} \quad (13)$$

1 T_{ref} and T_m are reference and melting temperature, respectively ($T_{ref} = 293$ K
 2 and $T_m = 873$ K). The expressions in the first, second and third sets of brack-
 3 ets represent respectively the effects of triaxiality (stress state), strain rate
 4 and temperature.

5 In the same way, to reflect the effect of temperature, the original Lou's
 6 ductile fracture criterion (Eq.8) is multiplied by the linear temperature term
 7 in the Johnson-Cook criterion leading to the following expression for the
 8 fracture strain:

$$\bar{\epsilon}_f(T) = \frac{C_3}{\left(\frac{2}{\sqrt{L^2+3}}\right)^{C_1} \left(\frac{\langle 1+3\eta \rangle}{2}\right)^{C_2}} (1 + C_4 T^*) \quad (14)$$

9 Accordingly, the extended form of Lou's ductile fracture criterion is ex-
 10 pressed as:

$$D(T) = \frac{1}{C_3(1 + C_4 T^*)} \int_0^{\bar{\epsilon}_f} \left(\frac{2}{\sqrt{L^2+3}}\right)^{C_1} \left(\frac{\langle 1+3\eta \rangle}{2}\right)^{C_2} d\bar{\epsilon}_p \quad (15)$$

11 It should be noted that the strain rate dependence of fracture strain is
 12 not considered in this work. For that reason the linear strain rate term in the
 13 Johnson-Cook model is omitted in the extended form of Lou's DF criterion.
 14 It should be also recalled that to avoid the influence of strain rate on ductile
 15 fracture, the cross head speed in fracture tests is optimized to ensure the
 16 same strain rate for all fracture specimens.

17 The extended form of Lou's ductile fracture criterion involves three pa-
 18 rameters (C_1 , C_2 and C_3) and an additional one C_4 to include the influence

1 of temperature. Since the material used in this work exhibits a weak Lode
 2 parameter dependency, the value of C_1 is set equal to zero. The remaining
 3 parameters are identified by using the procedure described in 2.4.2 based
 4 on all post-processed tests (i.e. $N_s = 4 \times 2 \times 3 = 24$). The identified values are
 5 $C_2 = 0.201$, $C_3 = 0.573$ and $C_4 = 1.321$.

6 The fracture loci obtained with this calibrated model are superimposed
 7 in Fig.14 in the space of stress triaxiality and equivalent plastic strain. It
 8 is observed that the fracture loci obtained with the extended form of DF
 9 criterion is close to the original form especially at elevated temperatures.
 10 Tab 8 shows the prediction error for the extended DF calibration at room
 11 and elevated temperatures.

		NR5	NR15	HR4	SH
RT	D_{pred}	0.873	1.035	1.032	1.006
	error (%)	12.73	3.54	3.17	0.61
	mean error (%)	5.01			
150 °	D_{pred}	0.999	1.002	1.015	1.014
	error (%)	0.09	0.23	1.53	1.39
	mean error (%)	0.81			
200 °	D_{pred}	1.069	0.968	0.927	0.990
	error (%)	6.90	3.17	7.28	0.93
	mean error (%)	4.57			

Table 8: Failure indicator D predicted in the critical element at the fracture stroke by the extended Lou's DF criterion and prediction errors (Eq.11).

12 It is found that the mean errors of the three tested temperatures are small
 13 and in the same range of those obtained by the original model. Indeed the

1 mean error for all tested temperature is found to be 3.46 %, which is very
2 close to the one obtained with the original model, about 3 %. These results
3 validate the extended form of DF criterion, which is very useful to predict
4 DF initiation under non-isothermal conditions usually encountered in warm
5 forming.

6 **4. Conclusions**

7 To figure out the relationship between forming limits, stress state and
8 temperature, ductile fracture prediction of AA6061-T6 aluminum alloy sheet
9 metal is investigated at room and elevated temperatures under a wide stress
10 state range. Specimens with different shapes are manufactured and de-
11 formed to achieve different stress states at RT, 150 °C and 200 °C. The hybrid
12 experimental-numerical approach is used to identify the fracture strain and
13 the corresponding stress state parameters (i.e. stress triaxiality and Lode
14 parameter). To accurately model the ductile fracture experiments, the FE
15 model is constrained by the real experimental boundary conditions and a
16 strain rate dependent anisotropic material model is employed. The com-
17 parison of the experimental and numerical results in terms of both force-
18 displacement curves and major strain distribution clearly demonstrates the
19 capability of the model to predict accurately the material response in differ-
20 ent loading conditions. Lou's DF criterion is used to consider the influence
21 of the stress parameters on the ductile fracture behavior. The DF is also
22 extended to add the dependence of temperature.

23 It is found out that the fracture strain is strongly dependent on the stress
24 state and temperature. The fracture strain may increase by more than a

1 factor of 1.7 when increasing the temperature from RT to 200 °C or by more
2 than a factor of 1.4 when decreasing the stress triaxiality value from 0.58 to
3 0.02.

4 It is also found out that the temperature effect on the stress triaxiality is
5 rather small and depends on the specimen geometry. The temperature has
6 less effect in case of the notched specimen NR15. However, the stress state
7 of the other specimens depends slightly on the testing temperature.

8 Also, results indicate that AA6061-T6 aluminum alloy exhibits a weak
9 Lode parameter dependency regardless of the tested temperature while a
10 moderate stress triaxiality sensitivity that decreases with the temperature
11 rise is noted. Moreover, the overall level of fracture strain increases by 31%
12 when temperature increases from RT to 200 °C.

13 The predictive capabilities of the original and extended criteria is eval-
14 uated using an error indicator. The mean error for all tested temperature
15 is found to be 3 % and 3.46 % for the original and extended criteria, respec-
16 tively. Therefore, it can be stated that Lou's DF criterion is well suited to
17 predict accurately ductile fracture at elevated temperatures under isothermal
18 conditions whereas the extended form of Lou's DF criterion is very useful to
19 predict DF initiation under non-isothermal conditions usually encountered
20 in warm forming.

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24

1 **Appendix A. Procedure to apply the real experimental boundary**
2 **conditions in Abaqus**

3 Fig.A shows the flowchart of the Python script used to define the bound-
4 ary conditions in Abaqus from the experimental DIC results. The experi-
5 mental local displacement filed on the surface of specimen is recorded during
6 test by the DIC software Aramis. Firstly, a reduced part of specimen with
7 a length $L = 45$ mm is defined in the specimen center. Then, a total of ten
8 points i are selected in the upper and lower sides of the reduced part of
9 specimen. The coordinates (X_i, Y_i, Z_i) of the selected points are extracted
10 at each time stage t . The initial coordinates (X_{i0}, Y_{i0}) determined at the
11 first stage ($t = t_0$) are used to define the 2D initial mesh in Abaqus that was
12 then extruded in the thickness direction to give the 3D mesh. Finally, the
13 experimental displacements calculated at the selected points (U_{xi}, U_{yi}, U_{zi})
14 are applied to the corresponding node sets to define the boundary conditions
15 in Abaqus.

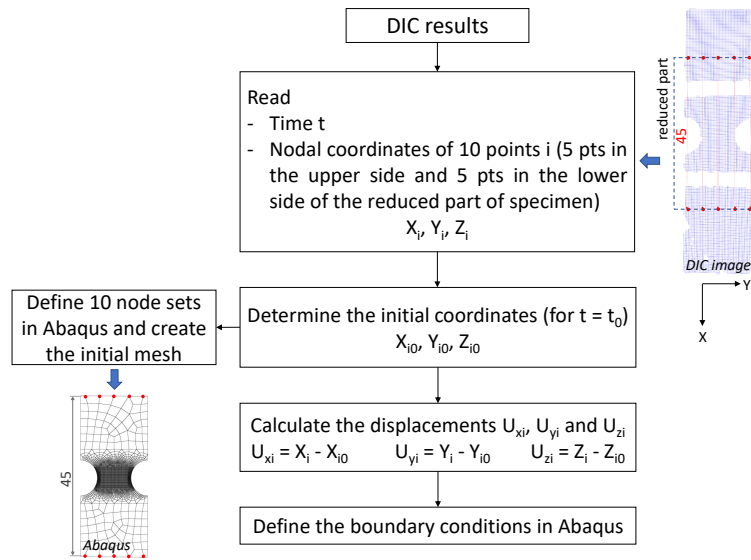


Figure A: Flowchart of the python script for applying the experimental local displacement, obtained from DIC results, to the FE model.

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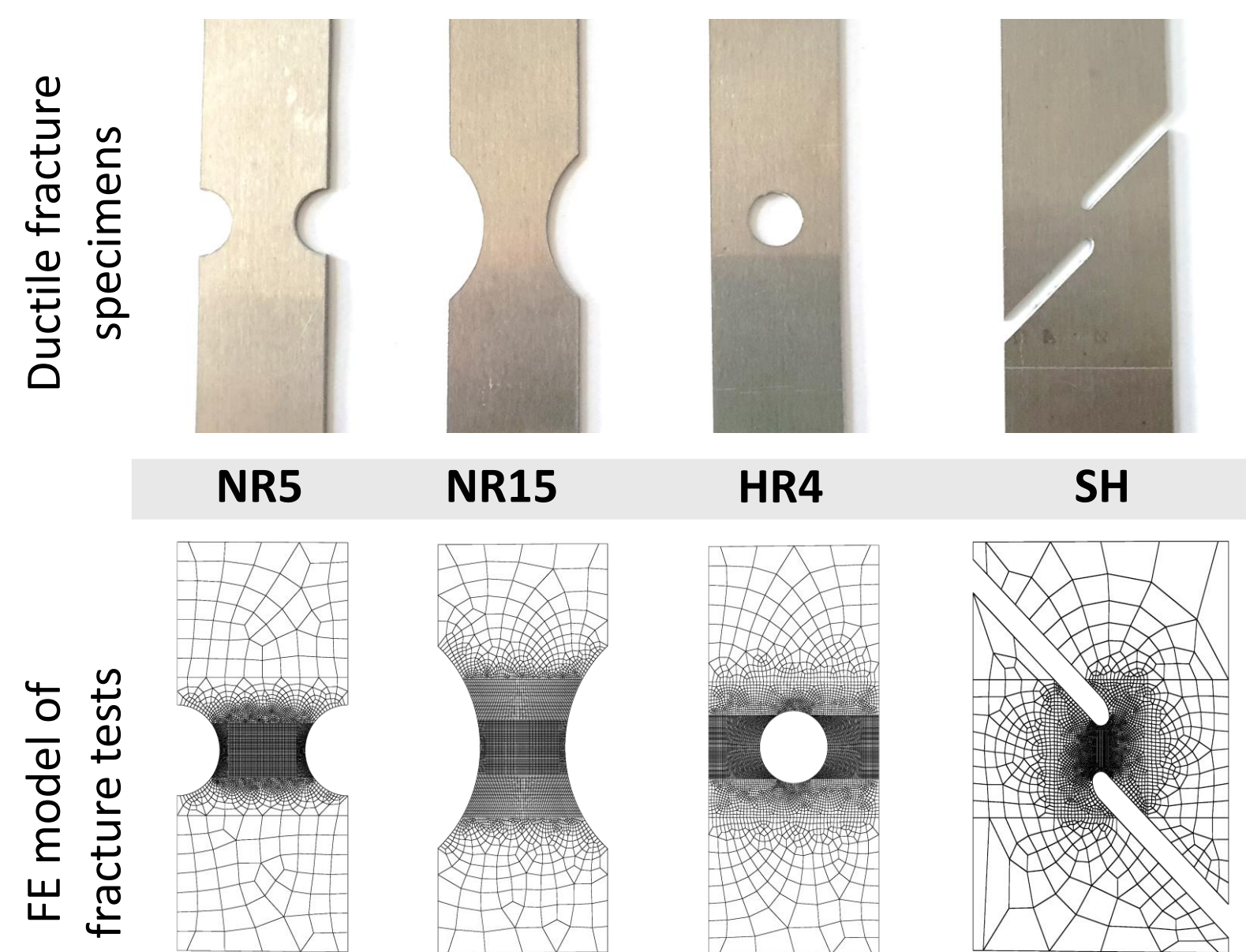
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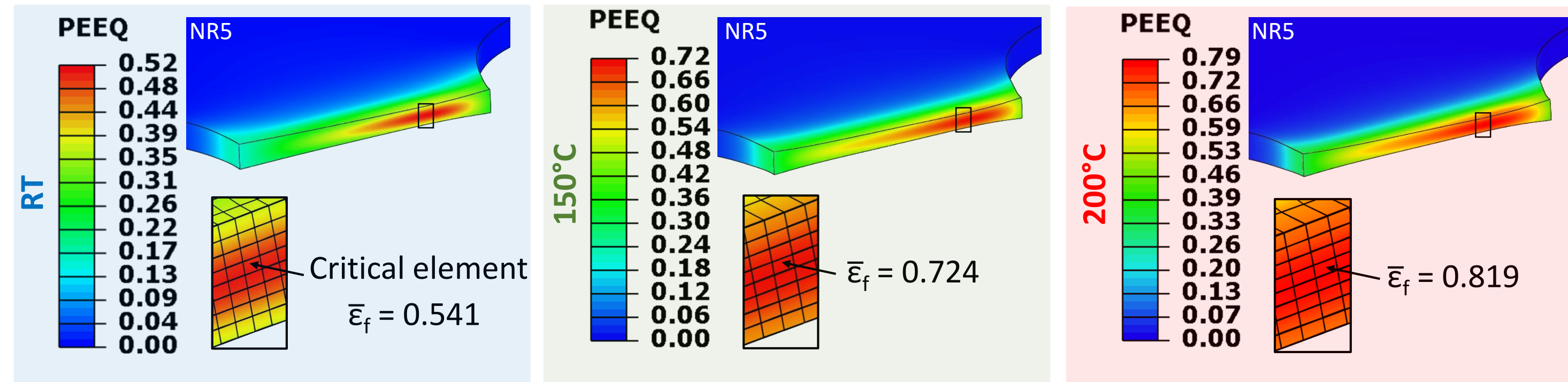
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Hybrid experimental-numerical approach



Determination of fracture strain at room and elevated temperatures



Fracture locus of AA6061-T6 at room and elevated temperatures

