EXPERIMENTAL AND NUMERICAL ANALYSIS OF THE FORMING LIMIT CURVE OF TYPE 304L AUSTENITIC STAINLESS STEEL

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Abstract
The experimental Forming Limit Curve (FLC) of 304L stainless steel has been determined by using the Marciniak punch test together with an optical method of direct correlation of digital images, which allows us to calculate the strain field under the punch. On the other hand, the Marciniak-Kuczynski (MK) approach has been used to predict the FLC. The calculations are developed by taking account of the effects of the martensitic phase transformation by means of the constitutive model developed by Iwamoto et al. for TRIP steels. The terminal stage of localisation is defined by using a fracture criterion in the neck, which is calibrated over the experimental observations.

1. INTRODUCTION
The concept of the Forming Limit Diagram (FLD) represents a tool of widespread practical use for assessing the limits to sheet metal ductility. The FLD is defined in the axes of minor (\(\varepsilon_2\)) and major (\(\varepsilon_1\)) principal strains in the plane of the sheet. The curve obtained by plotting the limit strains obtained for linear strain-paths is the Forming Limit Curve (FLC).

The most popular analysis for predicting the FLC is based on the model proposed by Marciniak and Kuczynski (1), who assumed the existence of an initial thickness defect across the sheet. Then, the limit strains result from the process of flow localisation in the defect. For certain materials, ductile fracture occurs without appreciable necking near equibiaxial tension, indicating that the evolution of damage becomes the controlling factor of ductility when the stress triaxiality increases. In several works the yield function proposed by Gurson (2) was used together with the M-K model to account for this effect. Another approach consists of using a fracture criterion to define the occurrence of failure in the defect (3). Then, reduced limit strains are predicted in the expansion range.

A limited number of studies have been devoted to the prediction of the limit strains in fully or partly austenitic steels exhibiting a martensitic transformation when they are plastically deformed at low temperature. This transformation is accompanied by a main specific effect, known as TRansformation Induced Plasticity (TRIP), or Greenwood-Johnson effect (4), which corresponds to the plastic deformation generated in the parent austenitic phase by the nucleation and growth of martensitic plates.

The FLC of type 304L stainless steel is determined in this work by means of the Marciniak test, where blanks of different widths are formed by the action of a flat-bottomed punch. Besides, an elastic-plastic flow localisation analysis is developed on the basis of the constitutive model proposed in a series of papers by Iwamoto et al. (5,6).
2. EXPERIMENTAL DETERMINATION OF THE FLC

2.1. Material
The experimental work was carried out on annealed 304L austenitic stainless steel sheets, 0.6 mm thick. The chemical composition is given in Table 1.

<table>
<thead>
<tr>
<th>Element</th>
<th>Cr</th>
<th>Ni</th>
<th>C</th>
<th>N</th>
<th>Mn</th>
<th>P</th>
<th>S</th>
<th>Si</th>
</tr>
</thead>
<tbody>
<tr>
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<td>8.05</td>
<td>0.042</td>
<td>0.05</td>
<td>1.27</td>
<td>0.028</td>
<td>0.001</td>
<td>0.55</td>
</tr>
</tbody>
</table>

Table 1. Chemical composition (weight percent) of 304L stainless steel.

2.2 Methodology of the Marciniak punch test
Sheets are deformed on a conventional tensile testing machine fitted with a Marciniak’s tool. This tool consists of a flat-bottomed cylindrical punch fixed on the machine and of a movable die and blank-holder. A drawbead is machined on the die and the blank-holder. This experimental device was developed by Brunet et al. (7).

The punch diameter is equal to 75 mm. The blanks are cut with different widths in order to generate strain-paths between uniaxial tension and biaxial stretching in the central part of the blank. A second blank having a central hole is placed between the punch and the specimen. For the smaller blank widths, i.e., in the draw region of the FLD, the second blank is simply cut in two pieces. A film of Teflon is inserted between the punch and the second blank. A clamping force of 200 kN is applied on the blank-holder at the set-up of the specimen. The displacement rate is equal to 10 mm/min.

2.3 Strain analysis
A mirror inclined at 45° restores the picture of the region of interest on a digital camera placed in front of the testing machine. Images are recorded before straining as well as at various levels of straining. At the beginning of the test, pictures are recorded every five seconds. In the terminal stages of the stamping process, the recording frequency is taken equal to three images per second to capture the onset of necking and fracture.

The strain measurement technique is based on a digital image correlation method developed by Brunet et al. (7). Each image of the deformed specimen is compared with the image of the undeformed surface using a grey level correlation coefficient. In order to perform more easily image correlation, a random grey level is needed on the samples before straining. This was obtained using two sprays of painting, consisting of a black speckle applied on a uniform white background.

The correlation technique is based on the search of a displacement field for a sub-image, or pattern. Correlation technique uses contrast between pixels of the same pattern. The in-plane displacement field is then used to calculate the strain field. The software (trade name Icasoft) allows us to obtain maps of the Green-Lagrange and/or Hencky strain measures.

2.4 Determination of limit strains
One point on the FLC at necking is the result of a particular post-processing, where the strain distribution is analysed along a line perpendicular to the fracture. The points relating to the local neck are eliminated, and a sinusoidal interpolation is made, where the limit strains at necking are defined at the apex of the interpolated curve. This procedure is similar to the one developed by Bragard (8) with a parabolic interpolation.
The limit strains at fracture were also evaluated. A first procedure consists of analysing the ultimate pictures where the image correlation method could be applied, so as to catch the in-plane ($\varepsilon_1, \varepsilon_2$) values at fracture. A cross checking was performed by means of a post-mortem analysis of the thickness profiles, giving the thickness strains $\varepsilon_3$ along a section normal to the fracture. The strain-paths at the fracture point were also determined.

3. NUMERICAL DETERMINATION OF THE FLC
The model developed by Iwamoto et al. (5,6) to account for the plastic behaviour of TRIP steels has been implemented in a program dedicated to the prediction of the FLC. This program is based on the localisation approach proposed by Marciniak and Kuczinski (1). The elastic-plastic behaviour of the material is taken into account in the calculations.

3.1 Material modelling

3.1.1. Inelastic strain-rates
The model assumes that inelastic straining in steels exhibiting strain-induced martensitic transformation can be described by the additive contributions of slip components, $\dot{\varepsilon}_{ij}^{\text{pslip}}$, accounting for dislocation glide in the biphasic material, and transformation components, ones describing the Greenwood-Johnson (G-J) effect, or shape change, $\dot{\varepsilon}_{ij}^{\text{pshape}}$, and the others the volumetric expansion associated with the transformation, $\dot{\varepsilon}_{ij}^{\text{pdilat}}$. Hence, the total inelastic strain-rate components $\dot{\varepsilon}_{ij}^P$ are expressed as:

$$\dot{\varepsilon}_{ij}^P = \dot{\varepsilon}_{ij}^{\text{pslip}} + \dot{\varepsilon}_{ij}^{\text{pshape}} + \dot{\varepsilon}_{ij}^{\text{pdilat}}$$  

(1)

The yield surface of the biphasic material is described by a yield function of the form:

$$f = F\left(\sigma_{ij}\right) - \bar{\sigma},$$

where $F\left(\sigma_{ij}\right)$ is a homogeneous function of degree one of the Cauchy stress components $\sigma_{ij}$ and $\bar{\sigma}$ is the effective yield stress of the biphasic material. The slip components are given by the normality rule:

$$\dot{\varepsilon}_{ij}^{\text{pslip}} = \dot{\varepsilon}^{\text{pslip}} \left(\partial F / \partial \sigma_{ij}\right)$$  

(2)

where $\dot{\varepsilon}^{\text{pslip}}$ is the work-conjugated effective plastic strain-rate for slip deformation. The strain-rate components accounting for the G-J effect are assumed to be collinear with the normal to the yield surface and proportional to the rate of transformation:

$$\dot{\varepsilon}_{ij}^{\text{pshape}} = R f_m \left(\partial F / \partial \sigma_{ij}\right)$$  

(3)

where $f_m$ is the volume fraction of martensite and $R$ is a material parameter. Finally, the dilatational strain-rate components are given by:

$$\dot{\varepsilon}_{ij}^{\text{pdilat}} = \frac{1}{3} \Delta v f_m \delta_{ij}$$  

(4)

where $\Delta v$ denotes the volumetric expansion resulting from the martensitic transformation.

3.1.2. Kinetics law for martensitic transformation
The kinetics law for strain-induced martensitic transformation follows the assumptions of the model of Olson and Cohen (9), which assumes that martensitic embryos form at shear-
band intersections. The increase in the volume fraction of shear-bands, $f_{sb}$, is described by the evolution law:

$$\frac{\dot{f}_{sb}}{1-f_{sb}} = a \dot{\varepsilon}_{a}^{\text{pslip}}$$

(5)

where $\alpha$ is a material parameter and $\dot{\varepsilon}_{a}^{\text{pslip}}$ is the effective plastic strain-rate in the austenitic phase. The number of operative nucleation sites is given by the number of shear-band intersections, multiplied by the probability $p$ that a shear-band intersection forms a martensitic embryo. The probability $p$ is defined by a Gaussian distribution function where the argument is the driving force of the martensitic transformation (10). The driving force $g$ is assumed to depend on temperature and stress-state, i.e.:

$$g = -T + g_{\Sigma}$$

(6)

where $T$ is the temperature, and $\Sigma = \sigma_{ii}/3\bar{\sigma}$ is the triaxiality factor; $\sigma_{ii}$ is the first invariant of the Cauchy stress tensor $\sigma$ and $\bar{\sigma}$ is the effective stress relating to the biphasic material.

3.1.3. Finite strains formulation
The constitutive equations are expressed for small elastic deformations. The total strain-increments are split in elastic and plastic parts, i.e. $\Delta \varepsilon_{ij} = \Delta \varepsilon_{ij}^{e} + \Delta \varepsilon_{ij}^{p}$. The elastic law is expressed in rate form using the Jaumann derivative of the Cauchy stress tensor $\sigma$. Thus, the stress and elastic strain-increments in the corotational frame are related by:

$$\Delta \sigma_{ij} = C_{ijkl} \Delta \varepsilon_{kl}$$

(7)

where $C_{ijkl}$ are the components of the elasticity tensor. The plastic spin is neglected. The measure of strains is chosen to be the integral of strain-rates in the corotational frame.

3.1.4. Tangent stiffness tensor
The calculations for elastic-plastic flow localisation require the determination of the tangent stiffness components $L_{ijkl}$ relating the stress increments $\Delta \sigma_{ij}$ to the total strain increments $\Delta \varepsilon_{kl}$ in the corotational frame, i.e.

$$\Delta \sigma_{ij} = L_{ijkl} \Delta \varepsilon_{kl}$$

(8)

An elastic predictor, plastic corrector integration scheme is used. The procedure is identical to the one previously formulated for numerical calculations using the finite element method (11). In the present calculations, the equations have been specialised to the case of plane-stress conditions for application to the flow localisation analysis described below.

3.2. Flow localisation analysis

3.2.1. Formulation and solution of the localisation problem
In agreement with the M-K analysis, the sheet is assumed to contain an initial imperfection, or defect, in the form of a band of reduced thickness. The homogeneous region and the defective region are denoted by the superscripts $a$ and $b$, respectively. The analysis is restricted to the case of a pure biaxial stretching imposed in the homogeneous region. In the present calculations, the strain-rate ratio $\rho^{a} = \Delta \varepsilon_{ii}^{a}/\Delta \varepsilon_{11}^{a}$ is kept constant (linear strain-paths). The initial conditions are defined by the initial thickness values, $h_{0}^{a}$ and $h_{0}^{b}$. The ratio $f = h_{0}^{b}/h_{0}^{a}$ defines the size of the defect. The equations that should be satisfied to ensure continuity between the two regions are the compatibility condition:
\[ \Delta \varepsilon^b_a = \Delta \varepsilon^a_a \]  
(9)

and the equilibrium conditions relating to the normal and tangential forces,
\[ \sigma^b_{mn} h^b = F_n ; \quad \sigma^b_{nt} h^b = F_t \]  
(10)

with \( F_n = \sigma^a_{mn} h^a \) and \( F_t = \sigma^a_{nt} h^a \); \( n \) and \( t \) are the normal and tangential directions to the band, respectively. For each increment, the condition imposed in region \( a \) is given by the pair of strain increments \((\Delta \varepsilon^a_1, \Delta \varepsilon^a_2)\). The kinematical conditions in region \( b \) are specified in the \((n, t)\) axes by \((\Delta \varepsilon^b_{mn}, \Delta \varepsilon^b_{nt}, \Delta \varepsilon^b_{nn})\) with \( \Delta \varepsilon^b_n = \Delta \varepsilon^a_n \). The iteration method aims at finding the \((\Delta \varepsilon^b_{mn}, \Delta \varepsilon^b_{nt})\) pair that satisfies the equilibrium Equations (10).

3.2.2. Definition of limit strains at necking

In the case when no fracture limit is assigned, the condition of flow localisation is defined by the attainment of a large value of the strain-rate ratio between the two regions. In practice, the calculations are halted when the effective strain-rate ratio \( \overline{\Delta \varepsilon^b_n} / \overline{\Delta \varepsilon^a_n} \) becomes larger than a prescribed value, taken equal to 10. This condition can be replaced by a criterion assuming the attainment of fracture in the defective region \( b \). The criterion proposed by Cockroft (12) to define the effective strain at fracture, \( \overline{\varepsilon_f} \), has been employed:
\[ \int_{\varepsilon_0}^{\overline{\varepsilon}} \sigma_1 \ d\varepsilon = C \]  
(11)

where \( \sigma_1 \) and \( \overline{\varepsilon} \) are the larger principal stress and effective strain in the neck (region \( b \)), respectively. The procedure for obtaining a point on the FLC is conducted as follows: (i) define the initial inclination \( \psi^a_0 \) of the defect, continue the incremental calculations until the condition of either localisation or fracture in the defective region is attained, (ii) define the limit strains at necking as the strain values \((\varepsilon^a_1, \varepsilon^a_2)\) when this condition is reached, (iii) repeat the calculations for various \( \psi^a_0 \)-values so as to find the minimum value of \( \varepsilon^a_1 \) and thus the \((\varepsilon^a_1, \varepsilon^a_2)\) pair defining the limit strains.

4. RESULTS AND DISCUSSION

4.1 Experimental

The blanks obtained after testing are shown on Figure 1. The blank widths vary from 60mm to 200mm, giving strain-paths between uniaxial tension and equibiaxial tension.

![Figure 1. Photographs of the blanks after punch stretching](image)

The experimental FLC is shown on Figure 2, together with the strain-paths recorded at the location where failure occurred. In spite of some scatter of the data points, a local maximum is clearly detected on the FLC at necking in the biaxial stretching range. The same behaviour was observed by Hecker et al. (13) and Talyan et al. (14). The strain values at fracture are also reported. The difference between limit strains at necking and at failure is small, particularly in the biaxial stretching range. This feature is consistent with
the observations made during the tests. Fracture occurs very suddenly, without any visible prior necking. The development of the neck could be detected on the consecutive pictures only near plane-strain tension.

![Figure 2. Strain-paths, and limit strains at necking (□) and at failure (O)](image)

A detailed analysis of the strain distributions at fracture was hence performed, using either the ultimate pictures where the image correlation method could be applied, or a post-mortem analysis of the specimens, allowing us to analyse the thickness profile along a section perpendicular to the fracture. The distribution of thickness strains perpendicular to fracture is shown on Figure 3. The measurements are made either with the image correlation method, by estimating the thickness strain from the measured surface strains with the assumption of incompressibility (Icasoft software), or by a direct measure of the thickness on an optical microscope (exp micro). A good agreement is obtained between the two methods, the more accurate being likely the image correlation method, because of the difficulty to obtain a precise measure of the thickness at the failure site. The absence of any indication of necking prior to fracture is particularly evident under biaxial stretching.
Figure 3. Thickness strains along a direction perpendicular to fracture. Blank width: 60 mm, uniaxial tension; 120 mm, plane strain tension; 200 mm, biaxial tension.

The optical micrographs of the profiles of fractured specimens are shown on Figure 4a for 3 typical strain paths. It is worth noting that, for all the strain-paths, the normal to the fractured surface is inclined at 45° from the plane of the sheet, indicating a shear-controlled fracture mechanism. The examination of fracture surfaces by scanning electron microscopy, Figure 4b, indicates that fracture is of the ductile dimple type.

Figure 4. (a) microscopic observation of the thickness profiles perpendicular to fracture. From left to right: blank width = 60 mm, uniaxial tension; 120 mm, plane strain tension; 200 mm, biaxial stretching. (b) Typical SEM fracture surface.

4.2. Comparison between experimental and calculated FLCs
The FLCs computed by assuming an initial thickness defect across the sheet, with an initial thickness ratio of 0.99, are shown on Figure 5. The limit strains at necking are calculated by assuming either neck growth controlled (NGC) limit strains or fracture controlled (FC)
limit strains. In that latter case, the calculations are halted when Cockroft’s fracture condition is attained in the defect. The limit value of Cockroft’s integral, i.e., constant $C$ in Equation (11), was adjusted to the experimental forming limit curve at fracture, Figure 2.

An excellent fit of the experimental FLC, shown in Figure 2, is obtained in Figure 5 by taking account of the attainment of fracture in the neck. Indeed, this observation is fully consistent with the experimental observation that fracture is the controlling factor defining the limit to ductility, at least in the biaxial stretching range.

5. CONCLUSIONS
The FLC of type 304L austenitic stainless steel is consistently predicted by means of a localisation approach where the occurrence of fracture in the neck is taken into account. The analysis is in agreement with the experimental observation of a very sudden occurrence of fracture without any indication of prior neck growth.

The localisation program also predicts the amount of martensite formed, which is around $f_m = 0.4-0.5$ when the experimentally-determined limit strains at fracture are reached. The possible role of martensite formation on the fracture of the specimens should be clarified. X-ray diffraction measurements of the fraction of martensite are in progress on the specimens that have been subjected to the Marciniak punch tests.

**Figure 5.** FLCs computed with the assumptions of either neck growth controlled (NGC) limit strains or fracture controlled (FC) limit strains. The curve corresponding to the attainment of Cockroft fracture condition in the neck is also reported.

**BIBLIOGRAPHY**