



# QUICK-PLASTIC FORMING: SIMILARITIES AND DIFFERENCES WITH SUPER-PLASTIC FORMING

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# Quick-Plastic forming: similarities and differences with Super-Plastic forming

R. BOISSIERE, S. TERZI, J.J. BLANDIN, L. SALVO

Grenoble-INP

Science and Engineering of Materials and Processes (SIMAP - GPM2)

CNRS, UJF, BP46, 38402 Saint-Martin d'Hères, France

## Abstract

*In order to extend the industrial use of SuperPlastic Forming (SPF), efforts have been devoted in the recent past to reduce forming times. The more spectacular result is probably the development of the so-called Quick Plastic Forming (QPF) technology which has been particularly investigated in the US automotive industry. QPF has been mainly studied in the case of Al-Mg alloy and to a much lesser extent in the case of magnesium alloys. In this presentation, some similarities and differences between QPF and SPF will be discussed. In particular, results will be shown concerning the respective mechanisms of deformation and fracture. A particular attention is paid to the sensitivity of the deformation and fracture mechanisms to temperature and strain rate of a wrought magnesium alloy. In this framework, data deduced from both mechanical characterisation (strain rate jump tests and tests at constant strain rate) and structural characterisation (deformed microstructures studied by microscopy and X-ray micro-tomography) will be presented.*

## Keywords

SPF, QPF, Forming, Al alloys, Mg alloys, deformation mechanisms, fracture mechanisms

## 1 INTRODUCTION

SuperPlastic Forming (SPF) of light alloys has been successfully used in aeronautics since more than 30 years. In the case of aluminium alloys, various alloys (2004, 5083 or 7475) have been formed with this technique. However, up to the beginning of the new millenary, the attempts for introducing SPF in the automotive industry have failed since only extremely small production volumes (i.e. typically for annual volumes between few hundred to exceptionally few thousand pieces) were concerned.

This difficulty could be attributed to two main causes: the cost of superplastic alloys and long forming times. Indeed, in order to get a high plastic stability and consequently to achieve large strains before failure, it is generally considered that grain boundary sliding (GBS) rather than dislocation creep (DC) must be activated as the main deformation mechanism. The interest to activate GBS comes from the fact that it is a mechanism of deformation associated with a high value of the strain rate sensitivity parameter  $m$  (i.e.  $m \approx 0.5$  or more for GBS whereas  $m$  is frequently close to 0.2 for DC) and the value of  $m$  is a key in the control of the ductility of a material since the higher the value of  $m$ , the higher the necking resistance. This rheology requirement has direct microstructural consequences: the activation of GBS requires to deal with alloys displaying fine grain sizes. In the case of aluminium alloys, complex (and consequently expensive) thermomechanical processing (TMP) was frequently required to fulfil this condition. Unfortunately, despite these efforts, the experimental window in which superplastic

properties can be reached are usually associated with relatively low strain rates, typically of about  $10^{-3} \text{ s}^{-1}$  or less. This means that the typical forming times in superplastic conditions must be considered in minutes rather than in seconds... Such forming times can be acceptable for the production of some complex components in aeronautics but remain too long for production volumes in automotive.

The reduction of forming time appears consequently as an important challenge. Two ways can be considered to reduce the forming times.

1. The first one is to increase the strain rate domain in which GBS can be preserved. This can be achieved by an additional reduction of the grain size. This way has been investigated thanks to the extensive experimental work dedicated to the superplastic characteristics of alloys having undergone severe plastic deformation. In such cases, structure sizes of about  $1 \mu\text{m}$  can be obtained and depending on the thermal stability of the alloys, superplastic deformation at strain rates up to  $0.1 \text{ s}^{-1}$  can be achieved. However, this kind of processing, if it can be industrially scaled up, will still increase the cost of the alloys.

2. The second one is to accept that GBS would be no longer the main mechanism of deformation. In other words, the idea is to find experimental conditions for which DC plays an important role (conversely to what occurs in “pure superplasticity”) but preserving significant values of  $m$  and consequently significant capacity of deformation. Such a view can give freedom on strain rates and consequently can result in significant reductions of forming times. This way has been successfully followed in the recent past by several US car makers [1,2] and the resulting forming process has been called Quick Plastic Forming (QPF).

The aim of this paper is twofold. Firstly, similarities and differences between QPF and SPF in terms of mechanisms of deformation and fracture will be discussed. Secondly, the associated behaviours of a 5083 aluminium alloy (Al-4.5Mg-0.7Mn, wt.%) and of a AZ31 magnesium alloy (Mg-3Al-1Zn, wt.%) will be compared in order to get information about the ability for the magnesium alloy to be a good candidate for QPF knowing the QPF properties of the Al 5083 alloy have been already reported in the literature.

## 2 QPF vs. SPF

### 2.1. DEFORMATION MECHANISMS

As already mentioned, the capacity to get a large plastic stability generally increases with the value of  $m$ . GBS is frequently associated with a value of  $m$  close to 0.5 whereas for DC,  $m$  is lower. For an alloy with a given microstructure (i.e. a given grain size) and deformed at a given temperature, a stress increase results generally in a continuous transition from GBS to DC. An important aspect is that these mechanisms have not the same dependency on the grain size: GBS is promoted by a grain size reduction whereas DC is roughly independent on grain size. For GBS, this grain size sensitivity depends also on the mechanism controlling GBS, volume diffusion or grain boundary diffusion.

Superplasticity will correspond to experimental conditions for which GBS will dominate. Consequently, at higher stresses (and/or at lower temperature), DC is expected to be the dominant mechanism. DC involves both glide and climb of dislocations and the lowest between these two mechanisms will control the deformation. In pure metals, climb is generally controlling and values of  $m \approx 0.2$  are frequently measured. In the case of alloys, dislocation glide can be the controlling mechanism, for instance due a solute drag effect. In such case,  $m$  increases to values between those measured when dislocation climb is controlling ( $m \approx 0.2$ ) and those measured in superplasticity ( $m \approx 0.5$ ). Indeed, when DC is controlled by viscous glide, values of about 0.3 are frequently measured. Keeping in mind that  $m$  affects the necking resistance, the possibility to identify an experimental domain in which viscous glide is the controlling mechanism is interesting in terms of forming capacity and forming time. In the case of aluminium alloys, QPF has been generally obtained for such experimental conditions.

In terms of microstructural changes, several differences can be pointed out between QPF and SPF. Firstly, in SPF conditions, due to the predominance of GBS, if the grains are initially equiaxed, this shape is maintained during the deformation. Conversely, in QPF, due to the predominance of DC, grains will deform and consequently grains elongated along the stress direction will be expected. However, one must keep in mind that during DC, dynamic recrystallisation processes can also occur, affecting the apparent shapes of the grains.

Another difference between QPF and SPF in terms of microstructural changes is related to crystallographic texture. For the same reasons than before (GBS in SPF and dislocation slip in DC), a randomisation of the texture (or maintain of an initially random texture) is expected during SPF whereas a progressive deformation texture is likely to appear in QPF. This difference can have consequences on the isotropy of the mechanical properties after forming.

## 2.2. FRACTURE MECHANISMS

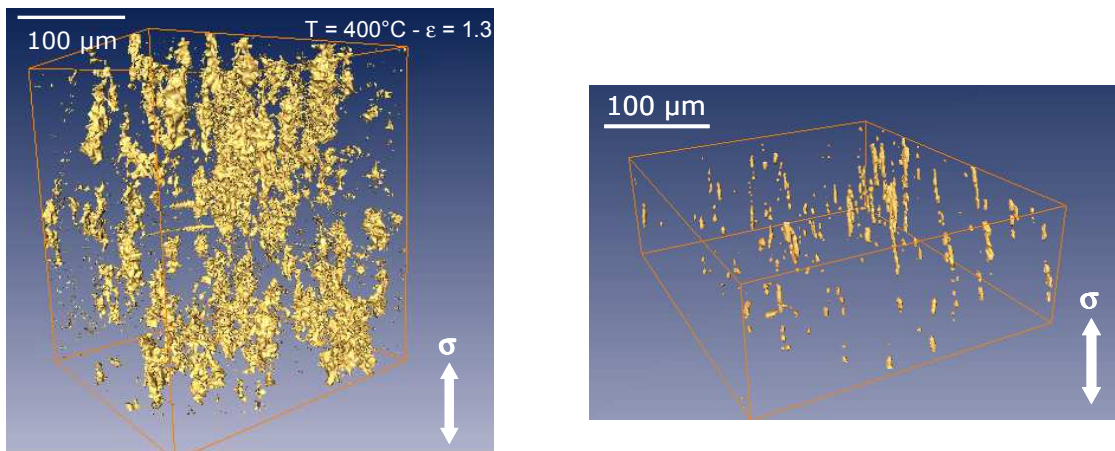
The  $m$ -value plays a key role in the resistance to necking by strain rate hardening. This effect will lead to more pronounced necks in samples fractured in QPF conditions. However, even if the value  $m$  is a necessary condition to get large deformability, it is not a sufficient one if strain induces damage. The point is that most single phase alloys (e.g. aluminium or magnesium alloys) are sensitive to strain induced damage by cavitation when they are deformed in superplastic conditions. In such conditions, fracture results generally from coalescence between cavities.

The variation with strain of the cavity volume fraction  $C_V$  during superplastic deformation is frequently described according to  $C_V = C_{V0} \exp(\eta_{app}\epsilon)$  with  $\eta_{app}$  an apparent growth parameter,  $C_{V0}$  a constant and  $\epsilon$  the macroscopic strain. If the variation of  $C_V$  with strain is supposed to result only from growth of individual cavities (which means that cavity nucleation is assumed to occur only in the early steps of strain), this relation can be associated to a particular mechanism of cavity growth (i.e. plasticity of the matrix around the cavity). In such conditions, the cavity growth parameter can be directly related to  $m$ . This means that the cavity growth rate is expected to depend on the deformation mechanisms and more precisely is supposed to decrease when  $m$  increases. Since the strain rate sensitivity parameter is higher for a viscous glide control than for climb control,

this means that cavities are expected to grow more slowly in the viscous glide regime (which corresponds to the QPF domain) than in the climb control regime.

For SPF, it must be emphasized that this mechanism of cavity growth by matrix plasticity does not take directly into account the role of GBS since it was initially developed for creep conditions. To overcome this difficulty, modifications of the law  $\eta(m)$  have been proposed in the past, introducing for instance a parameter varying from 0 for no GBS to 1 for freely sliding grain boundaries [3]. However, despite these improvements, such theoretical calculations frequently fail to fit the experimental cavitation behaviours which are observed in superplastic conditions.

As a result of the differences in deformation mechanisms, an important difference related to damage behaviours between QPF and SPF will be also related to the shapes of the cavities. In DC conditions (i.e. in QPF conditions), due to the intragranular deformation of the grains, cavities elongated along the stress direction are expected whereas in the superplastic regime, due to the predominance of GBS, cavity with particularly complex shapes will develop. Such a difference in cavity morphology is illustrated by figure 1 which shows 3D imaging of the population of cavities developed after superplastic deformation (fig. 1a) and DC deformation (fig. 1b) in the case of a fine grained magnesium alloy. Such 3D imaging has been obtained thanks to X-ray microtomography analyses. More detail concerning the use of this technique for characterising damage after high temperature deformation of metallic alloys have been published elsewhere [4-6].



**Figure 1**

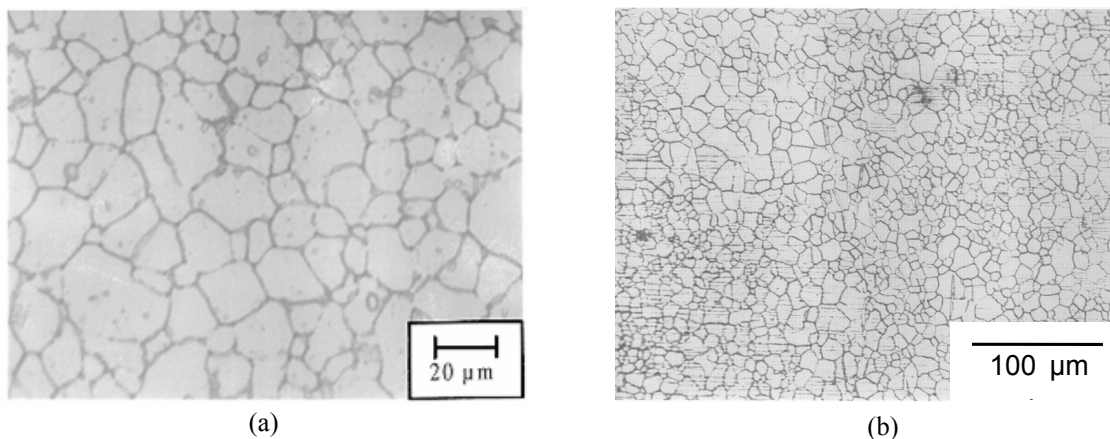
3D images showing typical populations of cavities after high temperature deformation of a magnesium alloy (fig. 1a) complex shapes of the cavities obtained in the superplastic domain (macroscopic tensile strain  $\approx 1.3$ ) (fig. 1b) elongated shapes of cavities developed when the alloy is deformed outside the superplastic domain (macroscopic tensile strain  $\approx 0.7$ )

Cavity nucleation will be also affected by the conditions of deformation. It has been established that the cavity nucleation rate in a given alloy is influenced by the flow stress level: the higher the stress, the higher the nucleation rate. Since QPF is associated with higher strain rates (and/or lower temperatures) than for SPF, higher stresses are measured. This means that higher cavity nucleation rates are expected in QPF conditions. This conclusion must however be taken with care since one very important parameter which controls the cavity nucleation rate, is also the characteristics of the population of particles in the alloy to be deformed. In the case of aluminium alloys, it has been for instance

demonstrated that the same alloy (e.g. Al 5083) could behave very differently in terms of damage behaviour depending on the particle characteristics. It is particularly true when the alloy is deformed in superplastic conditions and it is the reason why alloys with specific grades have been sometimes developed for SPF resulting in an additional increase of the cost.

### 3 AL 5083 *vs.* MG AZ31

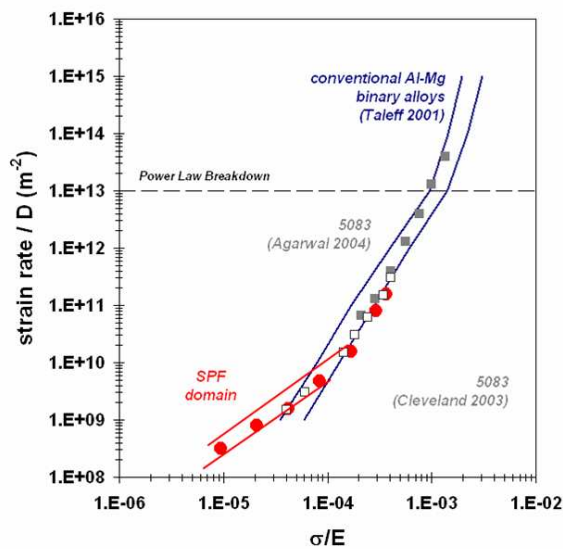
The studied Al 5083 alloy (Al-4.2Mg-0.8Mn-0.2Fe-0.1Cr, wt.%) was provided in the form of sheets of 2.5 mm thickness. In the as-received conditions, the alloy was not recrystallized and the recrystallization occurred during heating and holding at the temperature of deformation before deformation started. Additional data concerning the microstructure and the superplastic properties of this alloy can be found elsewhere [7]. Figure 2.a displays an optical micrograph of the microstructure observed in the sheet plane when deformation starts. A mean grain size in the range 10-15  $\mu\text{m}$  is measured and observations in the other planes have shown that the grains could be reasonably considered as equiaxed. The studied magnesium alloy was an AZ31 alloy (Mg-3Al-1Zn, wt.%), received as a 2 mm thick hot rolled sheet, with a mean recrystallized grain size in the same range as for the Al 5083 alloy. Figure 2-b displays an optical micrograph of the sheet plane of the as-received alloy [8,9] and observations in the other planes have again shown that the grains were equiaxed.



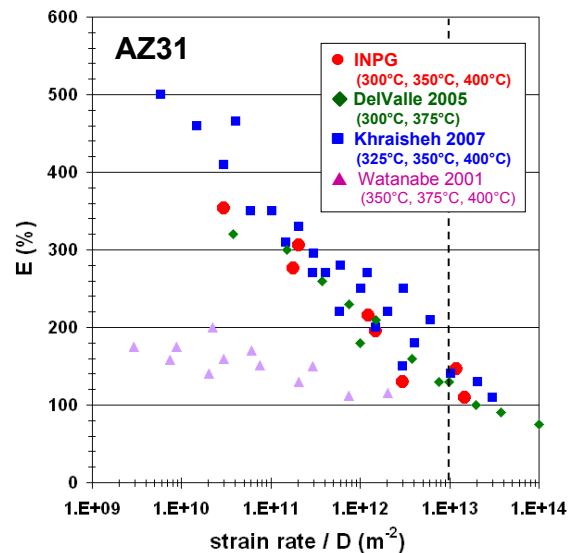
**Figure 2:** Optical micrographs of the studied alloys observed in the sheet planes  
 2.a. Al 5083 alloy after heating (i.e. when deformation starts)  
 2.b. as provided Mg AZ31 alloy

High temperature deformation of various Al-Mg alloys has been studied in particular by E. Taleff and co-workers [10-16]. In the case of fine grained Al 5083 alloys, superplastic properties are generally obtained between 450°C and 525°C and for strain rates roughly ranging from  $10^{-4} \text{ s}^{-1}$  to  $10^{-3} \text{ s}^{-1}$  [7,16]. In this domain,  $m$  values of about 0.5 have been frequently reported. When superplastic properties cannot be obtained, the Al-Mg alloys deform via a solute drag process in a large experimental domain. This rheology has been clearly identified in the case of binary Al-Mg alloys (with various Mg contents) but the addition of a ternary element, like manganese results in an apparent slight decrease of the  $m$  value, which consequently is lower than the expected value of 0.33 associated to a control by viscous glide. It is thanks to this large experimental domain in which the viscous glide of dislocation controls the deformation rate that significant strains to fracture

can be achieved, allowing the possibility to get QPF. Data collected for various Al-Mg alloys are summarised in figure 3. This figure displays the variation of the strain rate (compensated by the lattice diffusion coefficient) as a function of the flow stress (compensated by the Young's modulus). Three regimes can be roughly identified: the low stress domain is associated with superplastic properties when fine grained alloys are considered; at high stresses, low  $m$  values are measured and this domain is attributed to power law creep breakdown, as suggested by the value of  $\dot{\epsilon}/D \geq 10^{13} \text{ m}^{-2}$  (this value has been proposed as the limit over which power law creep is no longer dominant [17]). Between these two regions, an intermediate domain is observed for which significant elongations to fracture, typically higher than 200 % can be preserved. This domain is interpreted as the signature of a control of dislocation creep by viscous glide in Al-Mg alloys, due to the strong interaction between Mg solutes and the dislocations. Roughly speaking, it corresponds to the QPF conditions.



**Figure 3:** Variation with of the strain rate (compensated by the lattice diffusion coefficient) as a function of the flow stress (compensated by the Young's modulus) in the case of Al-Mg alloys.



**Figure 4:** Effect of strain rate (compensated by the lattice diffusion coefficient) on the elongation to fracture of various AZ31 alloys.

In the recent past, superplastic properties of magnesium alloys have received increasing attention [8,18,19]. In the case of fine grained AZ31 alloys, superplastic properties are generally obtained between 350°C and 400°C for strain rates ranging from  $10^{-4} \text{ s}^{-1}$  to  $10^{-3} \text{ s}^{-1}$ . In this domain, values of  $m$  of about 0.5 are frequently obtained and activation energies of about 90 kJ/mol are measured, suggesting that intergranular diffusion is predominant over lattice diffusion in the control of GBS for this alloy. For higher strain rates, the value of  $m$  is significantly reduced but the mechanisms of dislocation creep of the AZ31 alloy are still under debate: both a control by dislocation climb [18,20,21] or by a solute drag process [22,23] have been reported. These conclusions are generally drawn from the estimation of the  $m$  value, keeping in mind that an accurate measurement of  $m$  is never easy and that various techniques of measurement frequently result in various measured values.

The solute content is expected to play a key role in the emergence of solute drag as the controlling mechanism of DC in magnesium alloys. In the case of pure Mg, climb control

was reported for strain rates from  $10^{-6} \text{ s}^{-1}$  to  $10^{-3} \text{ s}^{-1}$  and  $T \leq 450^\circ\text{C}$  [24]. Conversely, for large solute amount, like in AZ91 alloys, rheologies associated with a value of  $m = 0.3$  were reported [25]. In the case of AZ31 alloys displaying superplastic properties (namely with fine grains), it seems that dislocation climb would be the rate controlling mechanism. This hypothesis is supported not only by the  $m$  values significantly lower than 0.3 but also by EBSD analyses carried out after deformation [18]. These analyses show that during deformation performed in the DC domain, a substructure is developed, suggesting that solutes have difficulty to drag dislocations or to inhibit intense recovery in the investigated alloy. This conclusion is also in agreement with the variations of the elongations to fracture with experimental conditions (i.e. strain rate compensated by the lattice diffusion coefficient), as illustrated by figure 4 for various AZ31 alloys [7,18,19,26]. On this figure are summarized data from alloys with fine grains (consequently displaying superplastic properties at low strain rates and high temperatures) and alloys with large grains which remain systematically in DC conditions whatever the applied conditions. For the fine grained alloys, a continuous decrease of the elongation to fracture is observed when  $\dot{\epsilon}/D$  increases and for the critical value  $\dot{\epsilon}/D = 10^{13} \text{ m}^{-2}$ , elongations between 100 and 150 % are measured. For the large grained alloys, elongations to fracture between 100 and 200 % are reached whatever the applied conditions. In particular, there is no apparent benefit to reduce the strain rate or to increase temperature in terms of deformability. In conclusion, the capacities of deformation of the AZ31 alloys appears lower than those displayed by the Al-Mg alloys, in particular when these alloys are deformed in a regime where DC is controlled by viscous glide. In other words, it appears more difficult to identify an experimental domain in which solute drag clearly controls deformation of AZ31 alloys.

A way to find such a domain could be to deal with Mg alloys containing a higher amount of solutes (e.g. AZ61 or ZK60 alloys) or if fine grained alloys are concerned, to form the components in a regime where GBS and climb controlled DC coexist. In this framework, it is interesting to remark that this coexistence can be preserved even at high strain rates due the intergranular diffusivity in magnesium alloys. Such maintain of the contribution of GBS in a large experimental interval can be observed in figure 4 since the increase in fracture strain when the strain rate is reduced is detected for values  $\dot{\epsilon}/D \approx 10^{13} \text{ m}^{-2}$ , whereas in the case of aluminium alloys, SPF rheologies were observed only for values of  $\dot{\epsilon}/D \approx 10^{10} \text{ m}^{-2}$  or less, as illustrated by the data shown in figure 3.

## 4 CONCLUSION

In the last years, Quick Plastic Forming has been successfully introduced as a forming technique of aluminium sheets in automotive industry. QPF derives from SPF in terms of capacity to reach at high temperature important strains during the forming process. Nevertheless, in the case of Al-Mg alloys, it is associated with a mechanism of deformation different from the predominant mechanism involved in superplasticity: viscous glide controlling dislocation creep for QPF and grain boundary sliding for SPF. This difference in mechanism induces several changes in microstructural variations during forming, in terms of texture, grain shape or damage behaviour. In the case of Mg alloys for which there is a growing interest in both aeronautic and automotive industries, the capacity to activate viscous glide controlling dislocation creep remains a relatively open question, in particular for the most studied Mg alloy (i.e. AZ31). Alloys with higher solute amounts would be probably better adapted to activate this mechanism. However, due to



particularly high grain boundary diffusivity, a significant contribution of GBS can be preserved in a large experimental domain, suggesting that components displaying significant strains could be formed quite rapidly.

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