Chapter 6

Mechanical performance of metallic glass nano-pillars in tension

Metallic glasses have drawn considerable attention due to the ability to resist high mechanical stresses; however, the lack of ductility remains the Achilles heel in the applicability to materials engineering and technology. This Chapter presents rather unique tensile experiments which were performed on Al$_{86}$Ni$_9$Y$_5$ and Zr$_{61.8}$Cu$_{18}$Ni$_{10.2}$Al$_{10}$ MG at nanoscale with diameters ranging between 100 and 500 nm. The in situ TEM observations were synchronized with picoindentation data and revealed a transition from brittleness on larger sizes to a plasticity performance while downsizing to hundreds of nanometers. Strain hardening effects were also observed. Our tensile experiments reveal the deformation behavior of MGs at nanoscale and address several relevant questions about the nature of intrinsic and extrinsic size effects at this length scale.

6.1 Introduction

Metallic glasses (MGs) have been widely studied as a new class of advanced materials owing to their desirable properties, such as outstanding yield strength and fracture strength, large elastic strain, and superior wear and corrosion resistance. Since the first publication by Paul Duwez and collaborators in the Sixties, metallic glasses have attracted considerable interest. However, their main drawback is the catastrophic and instantaneous brittle failure under tensile loading, originating from severe plastic-strain localization in a narrow region called a shear band.$^1$ As a consequence, it is of fundamental scientific and engineering interests to find out whether and how, the strength of MG alters when the physical dimensions are becoming smaller.$^2$ Also, in terms of fundamentals of plasticity, the way in which the specimen dimensions affect the ‘intrinsic’ strength and plasticity of amorphous metals is an issue of scientific relevance.$^3,^4$

For bulk metallic glasses, the plastic deformation is thought being localized in a narrow region of localized shear. Following micro-compression experiments on crystalline materials, several research groups have carried out
similar compression experiments on metallic glassy micropillars and reported a correlation between reduced size and several mechanical properties: maximum plastic strain before failure, yield strength and deformation mode. Nonetheless, several papers are inconsistent, sometimes due to lack of sufficient accurate experimental data and lack of statistics. Also, the stress-strain response of micro- and nanopillars during compression can be characterized by significant stochastic effects, both in crystalline and in glassy pillars. Another reason for the lack of agreement is the imperfect specimen geometry, which is, tapering and buckling of cylindrical pillars in compression tests.

In tension the plastic elongation of MGs is usually very small, i.e. virtually equal to zero. However, a higher density of shear transformation zones (STZs) may produce local shear transformations that can mediate a plastic flow without immediate catastrophic failure. Despite this potential and the fact that under specific conditions MGs can indeed exhibit plastic deformation, the problem with MGs is that plastic strains are commonly observed in extremely thin shear bands. The concentrated deformation in softer regions trigger rapid failure as soon as major shear banding sets in, leaving a little opportunity for an observable ductility in tension.

In contrast to non-crystalline materials, effects of specimen size on the apparent strength of crystalline materials have received a lot of attention, in particular under compressive loading. For single-crystal metals in the micrometer and submicron regime, a “smaller is stronger” trend has been established, because the smaller the specimen size, deformation mechanism becomes dislocation nucleation and starvation controlled (assuming no pre-existing dislocations), elevating the stress field required to cause yielding and sustain plastic flow. It is therefore intriguing whether an analogous “size effect” can also exist in MGs. A “the smaller the better” idea has arrived at the MGs research field. As such, small-volume MGs are especially attractive for applications such as in micro- and nano-electromechanical systems (MEMS and NEMS). A number of experiments have already been carried out in recent years to look into the possibility that specimen size influences the apparent strength of MG.

A variety of alloy systems have been examined, including MGs based on Pd, Zr, Fe and Mg. Unfortunately, the results have led to a controversial debate about deformation in compression as well as in tension. What happens if we decrease the geometrical size of the specimen close to or below the shear band size? Does the plastic flow mechanism change when its dimension goes
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down to the minimum thickness of a shear band, and if so, how? Therefore, it is of importance to carry out careful tests to elucidate thoroughly quantitative measurements whether nanometer sized MG specimens, at least those made using a focused ion beam (FIB), are intrinsically ductile under uniaxial compression as well as tension. In contrast to the previous chapters the present chapter concentrates on tensile experiments but a comparison with compression experimental results presented in previous chapters will also be made.

In-situ TEM tensile experiments were performed using a Hysitron picoindenter TEM holder (Hysitron Inc., Minneapolis, MN, USA) equipped on JEOL 2010F TEM, with a home-developed Al-alloy tip holder and W tensile tip. Al-alloy holder was designed and developed in order to achieve good calibration and to avoid undesirable mechanical noise due to over weight during tensile experiments. A tungsten tensile tip was designed and made by electrochemical polishing and afterwards milled by FIB to the desired shape (Fig. 6.1). The dimensions of the tip were chosen in order to make tensile experiments of nanopillars with diameter range from 100 to 500 nm (Fig. 6.1). The tensile stage has several features, which are particularly critical to the present study. It is integrated with a miniature capacitive load–displacement transducer permitting load and displacement measurements of high resolution (~0.3 µN in load, ~1 nm in displacement). The experiments have been run in displacement controlled mode with constant strain rate $\sim 10^{-2}$ s$^{-1}$.

![Fig. 6.1. TEM picture of FIB milled W tensile tip and Al$_{86}$Ni$_9$Y$_5$ MG nanopillar 250 nm in diameter](image)
6.2 Results and discussion

Tensile experiment on Al-based and Zr-based MGs

In particular we discuss tensile experiments of $\text{Al}_{86}\text{Ni}_{9}\text{Y}_{5}$ and $\text{Zr}_{61.8}\text{Cu}_{18}\text{Ni}_{10.2}\text{Al}_{10}$ MG pillars with diameters ranging from 100 to 500 nm. Upon testing specimens of larger diameter, the load increases proportionally up to yielding followed by brittle failure. The fracture surface shows a major shear phenomenon under 45 degrees, i.e. similar to the behavior of mm-sized MG specimens. The localization of the fracture could occur stochastically on the specimen surface over entire length, depending on the local concentration of smaller pre-shear which contribute to major shear band (Fig. 6.2).

![Fig. 6.2. Video frames recording the deformation of a $\phi$400 nm $\text{Al}_{86}\text{Ni}_{9}\text{Y}_{5}$ MG pillar in tension at displacement rate controlled mode. Grabbed video frames show the deformation structures before, during and after tension (a-d). True stress-strain curve (e).](image-url)
The deformation behavior alters in specimens with diameters around 250 nm. Upon tension, specimens show areas with a higher concentration of shear deformation (Fig. 6.3(a-d)). However, deformation results in catastrophic failure, as these areas interact with each other. The stress-strain curve (Fig. 6.3e) does not show any visible stress drops since the local deformation might occur at several positions on the specimen surface simultaneously and the picoindenter cannot resolve those when brittle failure occurs. An intriguing effect for these specimens is significant strain hardening (shown on Fig. 6.3e and Fig. 6.4e) that represents the size dependence in tension at this scale.

Fig. 6.3. Video frames recording the deformation of a ø220 nm Al_{86}Ni_{9}Y_{5} MG pillar in tension at displacement rate controlled mode. Grabbed video frames show the deformation structures before, during and after tension (a-d). True stress-strain curve (e).
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The localization of shear bands is different in larger pillars. The latter show more localized areas with shear bands, which propagate to one major shear band. The ultimate stress in smaller specimens is higher in comparison with larger diameter specimens. An increase of true strain up to 8% was observed for this diameter range of MG pillar specimens (Fig. 6.3e). The smallest specimen diameter in the range down to 100 nm shows more plasticity, necking and an increased strain hardening effect (Fig. 6.4e). Fracture behavior is similar to plastic crystalline materials, i.e. without a 45 degree fracture plane as occurs in larger specimens (Fig. 6.4(a-d)). The plastic strain reached is around 8%. The most important observation of tensile experiments is that no clear size effect is seen on the yield stress but a clear increase in the ultimate strength is observed with deceasing the diameter of MG pillars (Figs. 6.2e – 6.4e).

![Video frames recording the deformation of a ø100 nm Al\textsubscript{86}Ni\textsubscript{9}Y\textsubscript{5} MG pillar in tension at displacement rate controlled mode. Grabbed video frames show the deformation structures before, during and after tension (a-d). True stress-strain curve (e).](image)

**Fig. 6.4.** Video frames recording the deformation of a ø100 nm Al\textsubscript{86}Ni\textsubscript{9}Y\textsubscript{5} MG pillar in tension at displacement rate controlled mode. Grabbed video frames show the deformation structures before, during and after tension (a-d). True stress-strain curve (e).
According to our previous results (see Section 4.3.3) the MG composition has a large influence on the change in deformation behavior at the nanoscale. Therefore tensile experiments of Zr$_{61.8}$Cu$_{18}$Ni$_{10.2}$Al$_{10}$ MG have been performed in order to compare with and also show strain hardening effects observed in Al$_{86}$Ni$_{9}$Y$_{5}$ MG.

At larger diameters and before the transition threshold of Zr$_{61.8}$Cu$_{18}$Ni$_{10.2}$Al$_{10}$ MG, the specimens undergo brittle fracture after reaching a yield point. Fig. 6.5(a-d) shows TEM video frames of a pillar 350 nm in diameter before, during yielding and after brittle fracture. The yield stress was about 1050 MPa (Fig. 6.5e)

![Image of TEM video frames](image)

**Fig.6.5.** Video frames recording the deformation of a ø350 nm Zr$_{61.8}$Cu$_{18}$Ni$_{10.2}$Al$_{10}$ pillar in tension at displacement rate controlled mode. Grabbed video frames show the deformation structures before, during and after tension (a-d). True stress-strain curve (e).
Upon decreasing diameter $\text{Zr}_{61.8}\text{Cu}_{18}\text{Ni}_{10.2}\text{Al}_{10}$ specimens show up to 15% (relatively to the yield point) a strain hardening phenomenon before brittle fracture occurs (Fig. 6.6(a-d)). After yielding at $\sim 1000$ MPa specimens show strain hardening up to 1250 MPa and 4% of plastic strain (Fig. 6.6e). Again, the stress-strain curves of Zr-based MG pillars in the diameter range studied indicate no size effects on the yield stress but a clear increase in the ultimate strength (Figs. 6.5e – 6.7e).

![Image](image.png)

**Fig. 6.6.** Video frames recording the deformation of a $\varnothing 220$ nm $\text{Zr}_{61.8}\text{Cu}_{18}\text{Ni}_{10.2}\text{Al}_{10}$ MG pillar in tension at displacement rate controlled mode. Grabbed video frames show the deformation structures before, during and after tension (a-d). True stress-strain curve (e).
Tensile specimens with the diameters smaller than 120 nm show work hardening and after yielding, they reach an ultimate tensile strength up to 25% and 7% of plastic strain after yielding.

The most intriguing behavior in tension was observed in Al$_{86}$Ni$_{9}$Y$_{5}$ specimens smaller than 120 nm in diameter. Notwithstanding that larger pillars show failure via localized shear bands after elastic loading, smaller sized specimens, below 120 nm, show a rather ductile behavior: plasticity, work hardening, ultimate tensile strengthening and necking. Plastic deformation was also confirmed by a rather homogeneous decrease in specimen diameter (up to 4%) over the full length.

![Video frames recording the deformation of a ø120 nm Zr$_{61.8}$Cu$_{18}$Ni$_{10.2}$Al$_{10}$ MG pillar in tension at displacement rate controlled mode.](image)

Grabbed video frames show the deformation structures before, during and after tension (a-d). True stress-strain curve (e).
The plasticity that was achieved is about 8%, which is a significant value for this type of MG composition. The yield stress reaches 1000 MPa for the smallest pillar diameter which is similar and consistent with larger diameters, however strain hardening effect is significantly increasing upon decreasing specimen size (Fig. 6.6e and Fig. 6.7e).

Indeed, under compression we have noted intrinsic and extrinsic size effects in the experiments on the Al$_{86}$Ni$_9$Y$_5$ and Zr$_{61.8}$Cu$_{18}$Ni$_{10.2}$Al$_{10}$ (see previous chapters) and importantly the current tensile experiments allow to avoid some artifacts (tapering, shear band formation due to friction).

The phenomena of size effects in MGs at nanoscale can be explained in following manner. In order to generate major shear the large energy and certain interaction volume is required. Large pillars ~ 500 nm have enough volume and thus stored elastic energy to generate major shear band similar to bulk mm-sized specimens. Our previous compression tests on specimens of about 350 nm in diameter (chapter 5) show that major shear bands occur by a collective behavior of a certain density of smaller ones. Upon decreasing the diameter to 120 nm the number of shear events increases; however, they do not generate a major shear band because of the limited interaction volume, which is even smaller than for compression tests.

**Compression versus tension**

The difference in the mechanical response depends on intrinsic and extrinsic properties of the material investigated. In this section the comparison between compression and tensile experiments of MGs at the nanoscale will be discussed. Even though some bulk metallic glasses can exhibit compressive strains of over 25% [12], they still fracture abruptly under uniaxial tensile stresses. This fact indicates that metallic glasses exhibit a tension-compression asymmetry as far as plasticity is concerned. The latter is usually interpreted by continuum mechanics, which neglects the atomic structure of materials. However, metallic glasses show an asymmetry that can be linked to the amorphous structure and pre-treatments.

The interesting deformation behavior of Zr$_{61.8}$Cu$_{18}$Ni$_{10.2}$Al$_{10}$ specimens with cross-sections 60 nm by 266 nm is presented in Fig. 6.8. Obviously the specimens with the same cross-section (square) 1.6×10$^5$ nm$^2$ did not show any plasticity at this chemical composition. The region at the bottom of the specimen Fig. 6.8(c-d) undergoes plastic behavior. This observation points at the influence of the rectangular cross section (i.e.o. squared), which is different for the compression experiments.
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Fig. 6.8. Video frames recording the deformation of rectangular specimen (with dimensions $60 \times 266$ nm) of Zr$_{61.8}$Cu$_{18}$Ni$_{10.2}$Al$_{10}$ MG in tension at displacement rate controlled mode. Grabbed video frames show the deformation structures before, during and after tension (a-h). True stress-strain curve (i).
Fig. 6.9. True stress-strain curve of Al$_{86}$Ni$_9$Y$_5$ MG specimen of ø 200 nm in tension at displacement rate controlled mode.

Fig. 6.9 shows that another specimen does not exhibit the usual tensile behavior, i.e. release of intermittent shear bands and hardening effects (Fig. 6.9). Major fracture for this specimen has occurred at different places where smaller shear bands are released. Some of the shear bands released at the surface (Fig. 6.9). This is consistent with our compression experiments (Fig. 6.10). With increasing volume under compression the interaction between STZs has a larger chance to occur which leads to the build up and propagation of larger shear bands.

The comparison of stress level in compression and tensile experiments for two different compositions Al$_{86}$Ni$_9$Y$_5$ and Zr$_{61.8}$Cu$_{18}$Ni$_{10.2}$Al$_{10}$ MGs is displayed in Fig. 6.11. At both compositions there is no size effect of the yield stress observed, i.e. the same in tension and in compression. However both compositions show strain hardening upon reducing the specimen size (Fig. 6.11 (black curves)) under tension. There are several differences between the materials: Al-based specimens harden more than Zr-based, almost up to the same ultimate strength as for the bulk.

Both the tensile and compressive stresses increase the number of interactions among STZs. The specimens possess the same apparent Young’s modulus under tension and compression. However, the experiments show that ultimate strength under tension is higher than under compression.
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Fig. 6.10. Top: TEM picture of Al$_{86}$Ni$_9$Y$_5$ MG specimen of ø 250 nm in compression at displacement rate controlled mode; below: True stress-strain curve.
In our explanation of the asymmetry between tensile and compression experiments the stress state of the starting material plays a predominant role. Generally speaking, in making comparisons between the mechanical performance of materials it is of utmost importance to know precisely the
residual stress state at the start of an experimental test, under compression as well as under tension. In this regards, the effect of ion irradiation caused by FIB in the preparation stage of specimens has to be considered.

Obviously, the influence of irradiation is increasing with decreasing size of the specimen, since the ratio $2d_{Ga}/d$ (where constant $d_{Ga}$ is the penetration depth of Ga ions) is increasing upon decreasing the diameter $d$. For the smaller diameters $2d_{Ga}/d$ is reaching values equal to $1/4$ or more and the outer layer will be under compression. Therefore it is important to include the influence of the FIB ion layer for the smaller specimens. The negative compressive stress of the outer layer will generate an offset for the positive applied stress under tension. The ultimate tensile stress will therefore increase upon decreasing diameter as experimentally observed. One would expect also an offset under compressive loading, i.e. a decreasing peak stress upon decreasing diameter. This was not really observed in practice because of the experimental difficulties to perform reliable compression tests on pillars with 1 free-end and diameters smaller than 100 nm. More experimental efforts are needed to clarify this point.

Fig. 6.11 shows a considerable difference between the UTS values of Al$_{86}$Ni$_9$Y$_5$ MG and Zr$_{61.8}$Cu$_{18}$Ni$_{10.2}$Al$_{10}$ MG upon decreasing size. The difference is about a factor of 2, i.e. much larger for Al$_{86}$Ni$_9$Y$_5$ and Zr$_{61.8}$Cu$_{18}$Ni$_{10.2}$Al$_{10}$, say roughly 1500 MPa for Zr$_{61.8}$Cu$_{18}$Ni$_{10.2}$Al$_{10}$ and 2500 MPa for Al$_{86}$Ni$_9$Y$_5$ at a diameter of 100 nm and the question is whether we can understand these differences in a more quantitative sense.

Our starting hypothesis is the mean projected range of defects induced by the FIB treatment that may affect the UTS. It is worth noting that thermal or charge-based effects are discarded here: in a transmission electron microscope thermal and charge conditions could surpass the FIB in intensity but no deformation was observed during static observations. Therefore the explanation is caused by the presence of defects introduced into the material during ion milling pre-treatment and the subsequent knock-on effects produced as ions traveled within the material.

Stopping Range of Ions in Matter (SRIM, version 2012.03,$^{16}$) simulations allow us to see the effects of ions impinging upon a solid target. A typical result for SRIM of a material like Al exposed to a perpendicular 30kV Ga ion beam shows a projected damage range of 20 nm. Depending on the material, each ion generates an approximately fixed number of defects, the shape of the ion penetration distribution is then the same as the shape of the
distribution of defects created through ionic knock-on effects. The defect
distribution induced can be modeled as symmetrical and triangular.\textsuperscript{17}

We take the stress $\sigma$ and strain $\varepsilon$ interrelated by linear elasticity and
we relate the stress formed in the defect-rich region to the strain via the
modulus $E$ and Poisson’s ratio $\nu$:

$$\sigma(z) = \frac{E}{1-\nu} \cdot \varepsilon(z)$$  \hspace{1cm} (6.1)

During ion exposure various types of defects will be formed. In a crystalline
case it is easier to identify these defects as vacancies, self-interstitials and Ga
defects. We can express the contribution of defects to strain as linear, and thus
take the sum of each type of defect of concentration $C_{d_i}(z) = n_i / n$ and
atomic volume $\Omega_i$:

$$\varepsilon_d = \frac{1}{3} \sum_i \frac{\Delta \Omega_i}{\Omega} \cdot C_{d_i}(z)$$  \hspace{1cm} (6.2)

where $n_i / n$ is the number of defects over number of all lattice sites, where $i$,
refers to the various defects like vacancies, interstitials and Ga$^+$ defects, and
$\Delta \Omega_i / \Omega$ refers to the relative volume change due to specific kind of defect $i$.
Obviously in an amorphous system the reference state is not crystalline but a
state of random disorderliness. Therefore the damage cascade will be less
localized due to less focused knock-on collisions as it will be in the crystalline
state.

To predict $C_{d_i}(z) = n_i / n$ we take the displacement energy term $E_d$ of a
metal in the Kinchin-Pease equation that relates accelerated ion energy $E_a$
to the amount of defects formed through $E_d$.

$$n_d = \frac{E_d}{2E_a}$$  \hspace{1cm} (6.3)

$E_d$ is mostly measured empirically, and a large amount of literature is
available on experiments to measure $E_d$ of various materials\textsuperscript{18}, e.g. using
displacement energies of 16eV for Al the approximate amount of defects
formed by a single ionic impact at 30kV is 1000. If we estimate the ion
emission parameter of the FIB apparatus used at $2.08 \cdot 10^7$ ions per second
to achieve an ionic current of 10pA, we can state that approximately
$2 \cdot 10^{10}$ defects in crystalline aluminium would be generated. At low to medium
dose the number of defects goes linear with the fluence.
With \( n_d \) taken from Eq. (6.3) and substituting its value into Eqs. 6.1 and 6.2 it seems rather trivial to predict the residual stress. However, there is at least one snake. The contribution of defects to the total strain/stress is assumed to be linear. The stress field of a single defect in an infinite linear elastic solid can be calculated based on the theory presented already 70 years ago by Love \(^{19} \) and to date it can be done through atomistics. The latter predicts a stress of 0.18 MPa of a single defect in crystalline Al based alloy on the Daw-Baskes Embedded Atom Method (EAM) \(^{20} \). This would lead to an unrealistic high stress state and you may ask yourself: what is wrong here and where is the snake? The answer lies in the actual distribution. For a uniform distribution of defects in a linear elastic medium, the material will expand/contract uniformly leading to a stress-free material (assuming no other boundary conditions and constraints, e.g. surfaces etc). For a linear elastic solid this was worked out by ‘Jock’ Eshelby \(^{21} \) assuming a homogeneous distribution and for a non-linear elastic solid surprisingly enough only recently. \(^{22} \) The effective \( n_d \) for a triangular distribution and including a free surface is only 1% of the number predicted by the Kinchin-Pease equation.

Now we are in the position to make a fair estimate of the residual stress and to check our hypothesis concerning the increase of UTS upon decreasing size as displayed in Fig. 6.11.

In the milling process it is reasonable to estimate that only 15% of the Ga ions will impinge the surface perpendicularly. Therefore the effective dose is \( 4 \times 10^{19} \) ions m\(^{-2} \) and based on Kinchin-Pease equation each incoming ion will generate for the stress build-up effectively: \( 0.015 \times 1000 \times 4 \times 10^{19} = 6 \times 10^{20} \) defects m\(^{-2} \) for Al\(_{86}\)Ni\(_9\)Y\(_5\) MG and \( 3 \times 10^{20} \) defects m\(^{-2} \) for Zr\(_{61.8}\)Cu\(_{18}\)Ni\(_{10.2}\)Al\(_{10}\) (melting points and cohesive energies of Zr and Al differ a factor of 2). With a projected range of 20 nm at 30 kV Ga\(^+\) and the external dimensions of the experimental pillars (diameter 100 nm, length 300nm) the effective strain build-up is 0.013 in Al\(_{86}\)Ni\(_9\)Y\(_5\) MG and 0.007 in Zr\(_{61.8}\)Cu\(_{18}\)Ni\(_{10.2}\)Al\(_{10}\) taking into account that the density of amorphous systems is about 0.9 of crystalline materials systems and a volumetric change of 12% being the average with respect to Ga.

Based on these data and the elastic moduli and Poisson’s ratio as listed in Table 4.1 (\( E = 57.10^9 \) Pa, \( v = 0.384 \) for Al\(_{86}\)Ni\(_9\)Y\(_5\); for Zr\(_{61.8}\)Cu\(_{18}\)Ni\(_{10.2}\)Al\(_{10}\) \( E = 84.10^9 \) Pa, \( v = 0.377 \)) we estimate an increase of the UTS in pillars with diameter of 100 nm due to a compressive stress of 2.6 GPa in Al\(_{86}\)Ni\(_9\)Y\(_5\) and 1.3 GPa in Zr\(_{61.8}\)Cu\(_{18}\)Ni\(_{10.2}\)Al\(_{10}\) which turns out to be in fair agreement with experiments (Fig. 6.11).
This analysis provides support to our hypothesis that residual stresses play an important role in the UTS value of smaller sized pillars (range 100 nm in diameter). As expected, basically because of the difference in bond strength, the effects of residual stresses due to pre-treatment are more visible in Al-based than in Zr-based MG and therefore the ion irradiation effect has the largest impact onto the mechanical performance of smaller sized Al-based specimens under tension.

Likewise hardening under tension is expected because the interacting volume of the STZs is actually decreasing even more rapidly upon decreasing size and in fact confined due to the (relatively) increasing volume fraction of negative compressive stress. Indeed, for both MGs the larger the $d_{Ga}/2d$ ratio, the larger the hardening effect.

### 6.3 Conclusions

In summary, in-situ tensile tests of $\text{Al}_{86}\text{Ni}_{9}\text{Y}_{5}$ and $\text{Zr}_{61.8}\text{Cu}_{18}\text{Ni}_{10.2}\text{Al}_{10}$ MG specimens with diameters ranging between 100 and 500 nm show a predominant inhomogeneous flow characterized by shear banding. However starting at specimens with 120 nm in diameter, $\text{Al}_{86}\text{Ni}_{9}\text{Y}_{5}$ MG exhibits a change in deformation behaviour from brittle fracture to plastic behaviour and inhomogeneous plasticity of necking.

The influence of chemical composition on the transition threshold in the tensile experiments was also observed, in accordance with compression tests. $\text{Zr}_{61.8}\text{Cu}_{18}\text{Ni}_{10.2}\text{Al}_{10}$ MG still shows plasticity around 60 nm, which is confirmed by the localized plastic necking. The unique increase in strength and strain hardening was observed for smaller sized specimens in tension which is different compared to compression experiments with the same MG compositions.

The observations described in this Chapter 6 confirm our predictions about volumetric size effect of MGs at nanoscale and can be explained by the competition between the nucleation and propagation of STZs at different stress fields and volumes. The study shows that the asymmetry in compression-tensile experiments is due to the residual stress state caused by the FIB pre-treatment.
6.4 References
