Intermetallics 50 (2014) 54–58

Contents lists available at ScienceDirect

Intermetallics

journal homepage: www.elsevier.com/locate/intermet

Mechanical properties of Al–Cu–Fe quasicrystalline and crystalline phases: An analogy

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ARTICLE INFO

Article history:
Received 12 October 2013
Received in revised form
22 January 2014
Accepted 12 February 2014
Available online 12 March 2014

Keywords:
B. Plastic deformation mechanisms
B. Mechanical properties
B. yield stress
B. Stress relaxation

ABSTRACT

The mechanical properties of the ω-Al_7Cu_2Fe crystalline phase have been investigated over a large temperature range (650–1000 K). Despite of its antinomic structure with the icosahedral Al–Cu–Fe quasicrystalline phase, *i.e.* periodic *vs* non-periodic, its mechanical properties are very similar to those of the quasicrystalline phase, which strongly suggest similar deformation mechanisms. Consequently, as for the quasicrystalline structure, we propose that dislocation climb might control the plastic deformation of the ω-phase. However, in the present case, the specificities of the quasicrystalline structure cannot be invoked to justify the predominance of dislocation climb, which questions the role of quasiperiodicity on dislocation mobility. We suggest that this deformation mode certainly results from specific non-planar extensions of the dislocation core.

1. Introduction

The discovery of quasicrystals (QCs), ordered phases with non-periodic structures [1], was recently awarded by the Nobel Prize in Chemistry to D. Shechtman in 2011. Very quickly QCs induced a great interest into the scientific community and quasiperiodic structures were described using mathematical method [2,3], while an appropriate indexation system was assembled for the icosahedral (i) QCs [4]. Despite of their peculiar atomic arrangements, resulting from the absence of translational periodicity, dislocations were theoretically predicted in i-QCs [5,6] and experimentally observed by transmission electron microscopy (TEM) [7–9]. Rapidly *post-mortem* [10–12] as well as *in-situ* [13] TEM observations established that high-temperature plastic deformation of Al–Pd–Mn QCs originates from dislocation movement. Although TEM observations did not provide similar evidences for i-Al–Cu–Fe QCs, it is currently admitted that dislocation movements also control the plasticity of this alloy [14,15]. Mechanical properties and deformation models, in terms of dislocation movements, were recently reviewed [16–19].

Several models based on elementary dislocation mechanisms were proposed to account for the mechanical properties of i-QCs. The models generally attempt to account for the high brittle-to-ductile transition temperatures (BDTT) and to reproduce the compression stress–strain curves, which consist of a yield point followed by strain softening only [20]. Dislocation mobility is supposed to be controlled by a friction stress, which decreases with increasing plastic strain [21]. Whatever shear or climb, crystals or QCs, dislocation motion implies short-range order destruction-recombination, which is not specific to QCs. The motion of a leading dislocation introduces in its wake some disorder in the QC structure, which facilitates the motion of the following ones. This description, initially proposed by Friedel [22] was later on modelled by Guyot and Canova [23] and further developed by Feuerbacher et al. [24].

It has however been emphasised that a shift parallel to the plane of motion, *i.e.* the glide plane for crystals, destroys some geometrical connections of icosahedral tiling by creating new tile forms, thus strongly inhibiting this deformation mode, while, on the contrary, dislocation climb preserves the tiling [25]. Thus, while a phason field is associated with the two deformation mechanisms, dislocation climb of edge dislocations preserves the tiling in certain specific planes, called worm lines in 2D, with the only formation of phason walls. This process should be therefore the easiest to plastically deform icosahedral QCs, when atomic diffusion is sufficiently efficient. In this context, recent *in-situ* and *post-mortem* TEM observations on Al–Pd–Mn single QCs have convincingly given evidences that dislocation motion results from a climb process [25,26]. Mompiou et al. [27] developed a quantitative dislocation climb

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http://dx.doi.org/10.1016/j.intermet.2014.02.004
0966-9795/

Table 1
Compositions and structural characteristics of the ω- and the i-Al–Cu–Fe samples.

Samples	ω-$Al_{69.3}Cu_{20.1}Fe_{10.6}$	i-$Al_{65}Cu_{20}Fe_{15}$ [35]	i-$Al_{63.6}Cu_{24.0}Fe_{12.4}$ [32]
Processing	Powder metallurgy	Powder metallurgy	Conventional casting
Mean grain size (μm)	2	–	20

model to account for the mechanical properties and TEM observations of i-Al–Pd–Mn single QC [28]. The model is based on previous qualitative descriptions proposed in the seventies by Edelin and Poirier [29] and Le Hazif et al. [30] for interpreting the high temperature deformation of Mg and Be single crystals oriented in such a way that dislocation glide was suppressed.

The aim of the present study is to highlight the strong similarities between the mechanical properties of the ω-Al_7Cu_2Fe crystalline phase and the i-Al–Cu–Fe quasicrystalline phase, which address the question of the role of the specificities of the icosahedral quasicrystalline structure on its plastic behaviour. This study reports results that were for the first time obtained on the mechanical properties of the ω-Al_7Cu_2Fe crystalline phase. They are described in more details in Ref. [31]. The results for the i-Al–Cu–Fe quasicrystalline phase were mainly extracted from Refs. [14,32–35]. It must be emphasised that the ω-Al_7Cu_2Fe crystalline phase, is not considered as an approximant phase of the i-Al–Cu–Fe quasicrystalline phase. Clearly, the ω-phase does not exhibit neither the local icosahedral order, nor the specific electron density of 1.8 electron/atom of the i-phase and its approximant phases [36]. Note that it has been experimentally shown that the Al–Cu–Fe quasicrystalline phase [15,37] and some of its approximants exhibit similar mechanical properties and, consequently, should have similar plastic deformation mechanisms, as also suggested by Klein et al. [38]. Therefore, all the specific features of the quasicrystalline phase (such as phasons, pseudo-Mackay clusters, friction stress related to quasiperiodicity, …) that have been used to account for their peculiar mechanical properties cannot be invoked for the ω-phase.

2. Experimental details

Polycrystalline ω-Al_7Cu_2Fe samples of the P4/*mnc* space group (a = 0.63 nm and c = 1.48 nm) were produced by spark plasma sintering using pure Al and quasicrystalline Al–Cu–Fe powders [39,40]. The Al–Cu–Fe quasicrystalline ingots used to produce the powders were kindly provided by N. Baluc at the EPFL (Switzerland). Details regarding processing route and microstructural characteristics of the as-synthetised ω-Al_7Cu_2Fe phase are given in Ref. [40]. The microstructural characteristics of the ω-Al–Cu–Fe samples are given in Table 1, together with those of the i-Al–Cu–Fe samples. Compression tests were performed on parallelepipedic samples, size 2 mm × 2 mm × 4 mm, at different temperatures at a nominal strain rate of 2.10^{-4} s^{-1}. Experimental conditions and complete results of the various mechanical tests performed on the ω-Al_7Cu_2Fe phase are presented elsewhere [31]. For better comparison, temperatures are normalised with respect to the peritectic temperatures of each phase, which are $T_p \sim 1020$ K for the ω-phase [41] and $T_p \sim 1120$ K for the QC phase [42]. Note that we use here an average value for the peritectic temperature of the QC phase, which slightly depends on the alloy composition (±40 K).

3. Results and discussion

The temperature dependences of the Young's moduli are shown in Fig. 1 for both the ω-phase [40] and the i-phase [35]. For the two phases, the temperature dependence of Young's moduli are very similar to that of a typical metal. The slope is nearly constant at low temperature and becomes steeper at high temperature. The more pronounced decrease for the ω-phase above 800 K, as compared to the i-phase, certainly results from the lower peritectic temperature of the former phase. Note also that the equal value measured at room temperature (RT) for both phases may be fortuitous, since large discrepancies have been reported in the literature for the i-phase [35,40,43].

Fig. 2 shows the normalised hardness H/H_0, where H_0 is the hardness measured at RT, of the ω-phase [39] and of the i-phase [33] as a function of the normalised temperature $T_n = T/T_p$, where T_p is the peritectic transition temperature [41,42]. H_0 values are high for the two alloys, H_0 = 7 GPa and H_0 = 8 GPa for the ω-phase and the i-phase, respectively. For the two phases, H/H_0 exhibits a similar temperature behaviour, which can be decomposed into two temperature regimes. A first temperature regime, ranging between 0.3 T_p and 0.6 T_p, where H/H_0 does not vary significantly and a second regime for temperature above 0.6 T_p, where H/H_0 exhibits a clear decrease with increasing temperature. The transition temperature between the two regimes can be assimilated to the brittle-

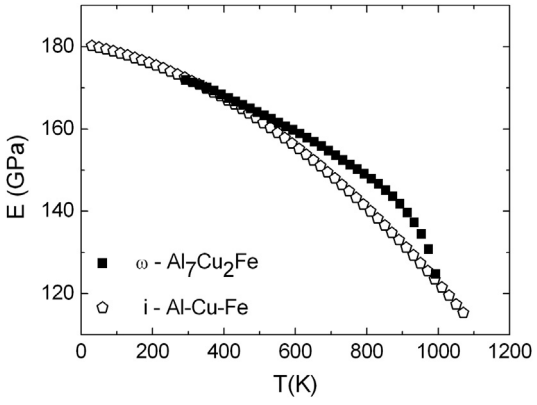

Fig. 1. Temperature dependence of Young's moduli for the ω- [40] and i-phases [35].

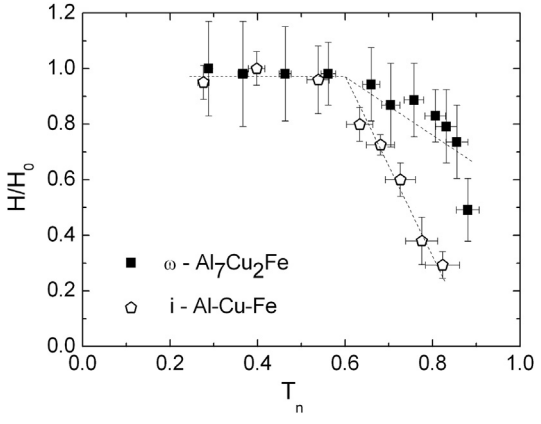

Fig. 2. Normalised hardness H/H_0 as a function of the normalised temperature, $T_n = T/T_p$, for the ω- [39] and i-phases [33] (see text for H_0 and T_p).