Relations between A_p temperature and deformation by constant stress in cooling in TiNi alloys

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The reverse martensitic transformation start temperature A_s of TiNi shape memory alloys can be elevated due to plastic deformation [1–7]. Lin *et al.* [1, 2] found that the A_s of TiNi alloys shifts to a higher value with increasing degree of cold rolling, and proposed that the deformed structures such as dislocations and vacancies contribute to the martensite stabilization. Piao *et al.* [3] observed that the A_s of TiNi alloys increases by tension at room temperature, and ascribed the effect to the releasing of the elastic strain energy by deformation. Zheng *et al.* [6] studied prestrained TiNi alloy fibers and the results confirmed that the release of the elastic energy is the main reason for the increase of A_s if the prestrain is within its recoverable limits. However, all these researches were made on the TiNi alloys deformed in the martensite state. If an external stress is added onto TiNi alloys in the parent phase, the alloys will be elongated upon following cooling process due to the presence of the external stress. But to the authors' knowledge, no research has been made on the reverse martensitic transformation behavior of the TiNi alloys deformed by constant stress in cooling.

It is well known that the self-accommodating martensite (SAM) will transform into the preferentially oriented martensite (POM, denoted as M1) as the wire is deformed in the martensite state. If the wire is deformed by constant stress in cooling from the temperature higher than A_f to the temperature lower than M_f, the martensite (denoted as M2) preferential to the deformation direction will be induced from the parent phase [8]. The differences of the stability mechanism between the M1 and the M2 will be discussed in this paper.

Commercial cold-drawn Ti-50.2 at.% Ni alloy wires with 0.48 mm in diameter were obtained from the General Research Institute for Non-Ferrous Metals, China. The wires were annealed at 873 K in vacuum for 1.8 ks followed by quenching into ice-water. The transformation behavior of the wires was determined by differential scanning calorimetry (DSC) using a TA INST 2910 DSC with a heating and cooling rate of 10 K/min. The martensitic transformation start and finish temperatures (M_s and M_f) and the reverse martensitic transformation start and finish temperatures (A_s and A_f) determined by DSC were 291, 280, 310, and 324 K, respectively.

The wires were deformed by the machine CSM-1, which can obtain the stress, strain, and temperature of the wires simultaneously. Firstly, the wires were heated to 373 K ($>A_f$), and then loaded to 27, 54, 162, 270, and 378 MPa, respectively. By the applied constant stress, the wires were elongated to different strains during cooling from 373 to 278 K ($<M_f$), due to the transformation from the parent to the martensite M2. Finally, the loads were removed and the strains of the wires were measured as 0.6, 1.7, 5.1, 6.4, and 7.5% for 27, 54, 162, 270, and 378 MPa, respectively. Obviously, these wires were composed of M2 martensite.

For comparison, some other wires were directly deformed by tension at 293 K (in the martensite state). After unloaded, the plastic strains of the wires were measured as 1.6, 3.4, 5.2, 6.4, and 7.7%. These wires were composed of M1 martensite.

The recovery strains of the deformed wires upon heating from 273 to 373 K were measured by CSM-1. The DSC specimens were cut from the deformed wires using a low-speed diamond saw.

The DSC heating curves of the wires deformed by constant stress in cooling are shown in Fig. 1. It is clearly seen that the endothermic peak of the reverse transformation of the M2 shifts to higher temperature with increasing deformation degree. The DSC heating curves of the M1 are consistent with those reported in previous study [5, 6], in which the endothermic peak of the reverse transformation of the M1 shifts to higher temperature with increasing deformation degree, thus omitted here. The peak temperature A_p (as shown in Fig. 1) of the DSC heating curve as a function of the deformation degree of two groups of the wires is shown in Fig. 2. It can be seen that the A_p of the M2 (curve 2) is lower than that of the M1 (curve 1) for the same deformation level.

The recovery strain ratio (the ratio of the recovery strain to the total deformation degree) as a function of the deformation degree of the wires is shown in Fig. 3. One can see that the recovery strain ratios of the two groups of the wires decrease with increasing deformation. The ratios of both decrease very slowly for the smaller deformation (<5%) and decrease rapidly for the larger deformation (>5%). For the same deformation degree, the recovery strain ratio of the M2 wires is lower than that of the M1 wires.

Fig. 4 shows the temperature A_p as a function of the residual strain of the wires. With increasing residual

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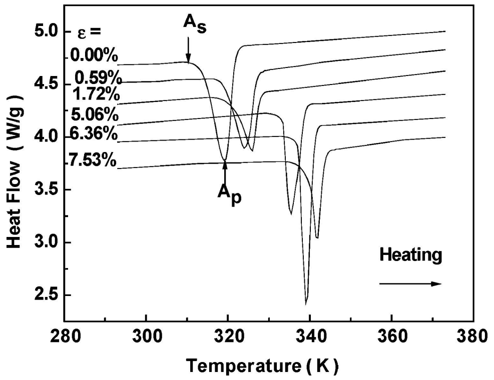

Figure 1 Heating DCS curves of the wires deformed by constant stress in cooling.

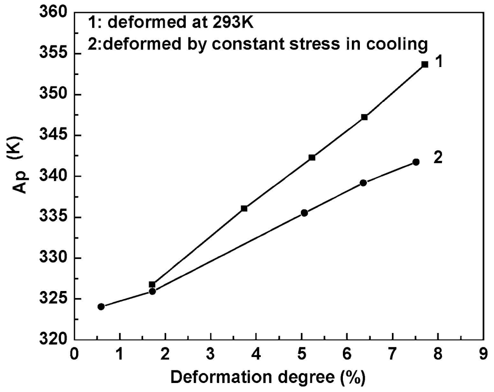

Figure 2 The peak temperature A_p of the DSC heating curve as a function of the deformation degree of the wires.

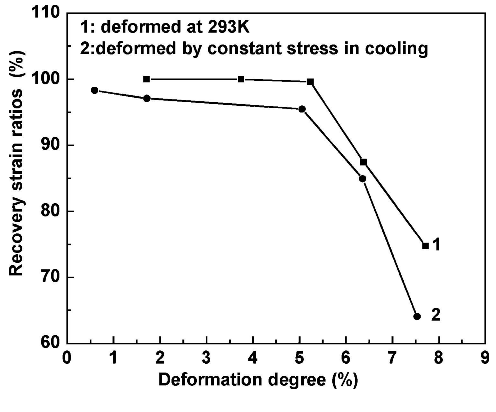

Figure 3 The recovery strain ratio as a function of the deformation degree of the wires.

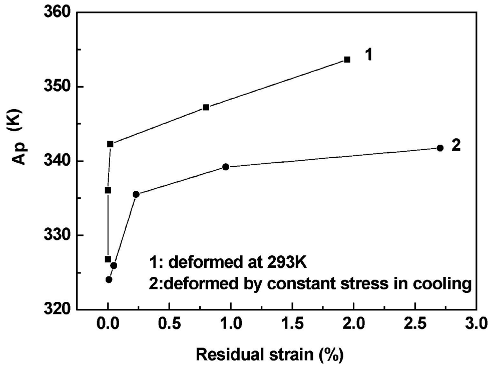

Figure 4 The temperature A_p as a function of the residual strain of the wires.

strain, the A_p of both increases rapidly for small residual strain, and quiet slowly for further increase in the residual strain. For the same residual strain, the A_p of the M2 (curve 2) is lower than that of the M1 (curve 1).

Lin *et al.* [1, 2] found that the A_s of TiNi alloys shifts to a higher value with increasing degree of deformation, and proposed that the dislocations and vacancies contribute to the martensite stabilization. Our experiment showed that the recovery strain ratio of the M2 wires is lower than that of the M1 wires for the same deformation degree, as shown in Fig. 3. A lower recovery strain ratio means a higher density of dislocations. Thus from Lin's viewpoint, the A_s of the M2 should be higher than that of the M1, which means that the A_p of the M2 should be higher than that of the M1. However, an opposite result is shown in Fig. 2. This phenomenon indicates that the dislocations may not be an important factor in the increase of the reverse transformation temperatures.

Zheng *et al.* [6] found that if the deformation of TiNi alloys is within its recoverable limits ($<8\%$), the plastic strain can be recovered upon heating with no or very little residual strain left but the A_s increases dramatically with increasing deformation, which is consistent with the results in the present study that the A_p increases with increasing deformation (see Fig. 2). Zheng *et al.* [6] proposed that the elastic strain energy is released during deformation ($<8\%$), then the A_s is elevated. However, it can be seen in Fig. 3 that when the deformation is 5% higher, the recovery ratios of the wires decrease rapidly with increasing deformation. It can be seen in Fig. 4 that when the residual strain is higher than a certain value (the deformation degree of about 5%), the A_p of the wires increase very slowly with increasing residual strain. Thus it can be induced that the elastic strain energy decreases rapidly to the minimum value while the deformation degree increases to 5% with almost no dislocations generated. When the deformation is higher than 5%, the amount of the dislocations increases rapidly, but the elastic strain energy does not change significantly with increasing deformation degree. As seen in Fig. 4, the decrease of the elastic strain energy is the main factor in the small residual strain range (deformation less than 5%) and the deformation-induced dislocation is the main factor in the large residual strain range (deformation higher than 5%) for the stabilization of martensite. However, the effect of the release of the elastic strain energy is more significant than the generation of dislocation on the increase of A_p.

The following is the summary of the present work. The A_p of the TiNi wires deformed by constant stress in cooling increases with increasing deformation. The release of the elastic strain energy and the deformation-induced dislocations both contribute to the increase of the A_p, but the effect of the former is more significant than that of the latter.

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