

THE MECHANICAL BEHAVIOUR OF A Cu-Zn-Al SHAPE MEMORY ALLOY WITH A SINGLE-INTERFACE TRANSFORMATION

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Abstract:

The stress-induced martensitic transformation with a single interface in a single crystal of the *Cu-Zn-Al* alloy is investigated. The transformation with entry two once more as for the usual $\beta \rightarrow 18R$ transformation starts as a martensitic band nucleation after the critical barrier surmounting. The thermoelastic equilibrium after the band nucleation is not reached. Therefore the subsequent martensitic deformation accompanies with a continuous stress-drop at a constant temperature (pseudoelasticity) or with a temperature increasing at a constant load (shape memory). The transformation has a very narrow hysteresis about $1 \div 2$ K.

Introduction

The martensitic transformation with a single interface is a case investigated by that no elastic energy should be saved and no transformation thermoelasticity is expected. The last means that the martensite and therefore martensitic deformation increases with the temperature decreasing or stress increasing.

But there is no clear definition what is a transformation with a single interface as an investigation object: microscopical growth of one of the more isolated, not interacted martensitic plate that is to macroscopical behaviour of the specimen integrated or real the nucleation and growth of the single mart-

ensit crystal that defines one to one the macroscopical characteristics martensit fraction, martensitic deformation etc.

The results of microscopical observations show either no dependence of the interface position change on the temperature [1] or a light slope of that [2], whereas macroscopical properties are not investigated. More investigations of the pseudoelastic behaviour corresponding to the transformations with a single interface shown the unloading of samples during the martensitic deformation after the critical stress [3-5].

It has been also shown theoretical from the analysis of a nucleation kinetics and a kinetical relation for the quasi-static stress induced martensitic trans-

formation with a single interface that the load drops after nucleation at a defined rates ratio of the loading and the interface motion [5].

On the other hand the thermodynamical treatment of martensitic multi-interface transformations and pseudoelasticity give a stress-strain line with a negative slope and a strain-temperature line with a positive slope that corresponds to an instable two-phase equilibrium [6,7].

Experimentally this line is found in partial cycles, whereas the martensitic deformation depends not on the load [7]. There is not a talk of thermoelasticity in this theory, because the elastic coherent interface energy is relevant to the hysteresis, and it follows that one is narrow for the transformation with a single interface.

In this paper are investigated the thermodynamical and kinetical characteristics of the stress-induced martensitic transformation, its hysteresis and corresponded shape-memory and pseudoelastic behaviour in the alloy, which is also investigated in the case of a multi-interface transformation [7].

Investigation of these transformation corresponded properties could be very interesting for the practical purposes because close temperature and stress intervalls are required for the certain applications of the memory elements.

Experimental techniques and results

Single crystals are grown by the Bridgman methode with argon poured in and sealed

quartz ampoules from a melt with the nominal composition *Cu- 16.0 at % Zn-16.0 at % Al* ($e/a = 1.48$).

The samples of square cross section of 2mm by 2mm, a length and thicker ends are prepared for tensile tests in an in-situ mashine with parallel transformation morphology observation in an optical microscop with a temperature cange 270-370 K.

Round discs of 0.5 mm thickness and 5mm diameter are used for calorimetrical measurements by by DSC-7.

Electrical resistivity is measured in specimens of cross section of 0.5 mm by 1 mm and length 70 mm by the standard four contacts method.

The transformation temperatures are measured by these both methods of about $M_S = 268K$, $M_f = 255K$, $A_S = 260K$ and $A_f = 271K$. The specimens temperature was regulated by a thermostat and registrated by the thermoelement PT-100, contacted with the specimen surface.

The gripps controlled elongation velocity is $10^{-6} \text{ m}\cdot\text{s}^{-1}$ by the stress-strain tests at a constant temperature. The dependences of the specimen elongation on the temperature, load and time are recorded by both experiment types.

A typical stress-strain (P-d) diagram in a tension-compression test at 285K is shown in Figure 1. This diagram is recorded after several cycles during that only one martensitic band has been induced.

The hysteresis in the both tension and compression sides is so small that in this scale by the high yield stress can not be resolved. Figure 1a shows a re-

scaled stress-strain curve by the tension; it has the stress-drop character.

The drop in stress level takes place after the critical barrier

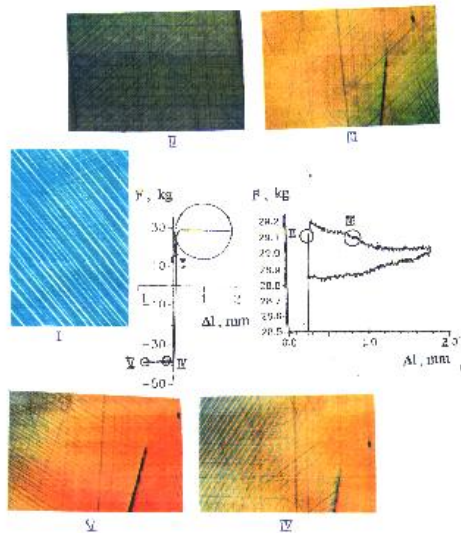


Fig.1. Tension and compression stress-strain curves at the temperature 285K; (a) rescaled tension curve surmounting and nucleation of a martensitic band of the about 1.5 mm critical width, as showed by microscopical observation (Fig. 2b).

But long before the band nucleation and yield stress the

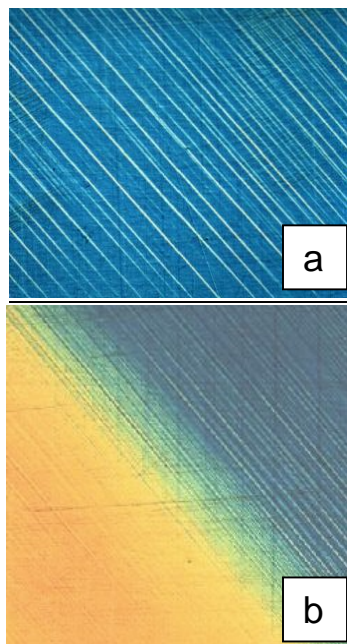


Fig. 2. Optical micrographs of the specimen surface at the load $P = 0$ (a) and $P = 80$ MPa (b). The test temperature of 300K

martensitic fine striations appear regular across a sample surface (Fig.2a).

Their density increases on the stress increasing without additional, inelastic deformation of specimens. It means, that these striations are martensitic twins. The density of twins reached at the yield stress on the nucleation place of about 50 mm^{-1} with a distances between the twins of about 10 mm^{-1} equaled their width.

The nucleated martensitic band has no sharp boundary. Its width covered about $0.3 \div 0.5$ mm. The martensitic deformation Δd within a band could be estimated as a relation of the measured elongation Δl to the band width b for the same time

$$\Delta d = \Delta l / b \approx 0.07 \quad (1)$$

This value remains approximately constant during the band growth through the interface motion and the corresponded specimen elongation.

The dependence of the yield stress on the temperature on tension side has a smaller slope than the one on compression side, whereas the transformation strain released on compression side is always smaller than the one on tension side (Fig. 3). The dependence of the critical stress of the martensitic twins appearing on the temperature has a slope of the two times smaller than one of the yield stress dependence.

The strain-temperature (d-T) behaviour at a constant load is fundamentally similar to the stress-strain one. An interface of the martensitic band moved spontaneous whereas external temperature and stress remain

constant. Figure 4 shows an isothermal martensitic deformation at the constant stress of 40 MPa after cooling or heating by the rate of $0.02 \text{ K}\cdot\text{min}^{-1}$ up to beginning transformation and elongation.

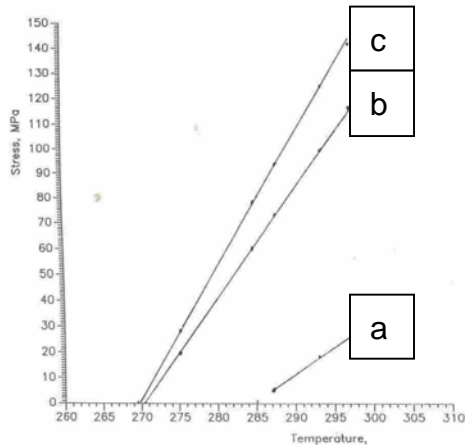


Fig.3. Dependence of the critical stress for the martensitic twins (a), yield stress on tension (b) and compression (inverse) (c) sides on the temperature

At this temperature the deformation increases at external isothermal conditions. The temperature hysteresis of about 2 K has been measured by as well partial as complete cycling.

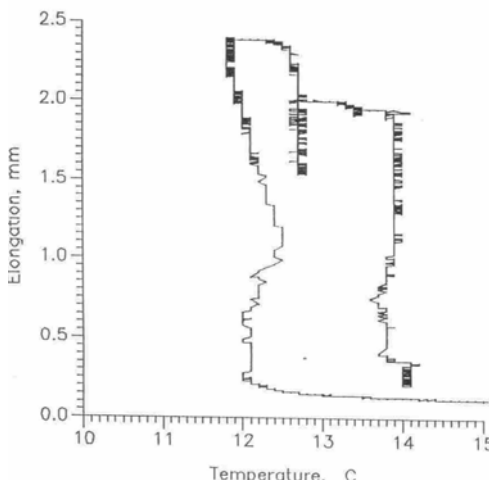


Fig.4. Strain-temperature curve at the constant stress 40 MPa

Within this hysteresis a metastable equilibrium of a single interface can be reached, where the martensitic deformation of the specimen could decrease or increase with the time. Exothermal and endothermal effects

of about 0.2 K at this isothermal conditions and about 2 K at adiabatic one (cooling rate $1 \text{ K}\cdot\text{min}^{-1}$).

Interface velocity is found from the elongation-time diagram for the nucleation within the range between the twin appearance and the yield, and growth. A small martensitic deformation is registered on the nucleation stage and its change has the thermoelastic character - increases on the cooling. The interface velocity of $(5\div 10)\cdot 10^{-3} \text{ mm}$ after band nucleation is measured. This value is much bigger than the grips controlled deformation rate by stress-strain experiments.

Discussion

The results show that by changing the temperature and by applying stresses two stage of the martensitic transformation with the same shear plane (the same Schmid factor) can be distinguished.

The temperature dependence of the both critical shear stresses is represented by right lines with the ratio of their slopes of approximately 1:2, which is known for the more different alloys and transformations, where the transformation kinetics is changed from the multi-interface to the single-interface transformation [3,4,8].

This stress-temperature dependences can be well explained by the Clausius-Clapeyron equation:

$$dP/dT = \Delta\eta/\Delta d \quad (2)$$

where $\Delta\eta$ and Δd are the transformation entropy and strain change. If the ratio $\Delta d_T/\Delta d_C$ of

about 1.3 of transformation strain change on tension and compression side would taken into account, the entropy change $\Delta\eta = (0.035 \pm 0.005) J/g \cdot K$ will be equal for the both.

The dependence of the critical stress of martensitic twins appearance on the temperature has the same slope as the $\beta \rightarrow 18R$ multi-interface transformation in this alloy [7] with the in eq.(1) calculated transformation strain of about 0.07 and transformation entropy of $(0.014 \pm 0.003) J/g \cdot K$.

A similar transformation kinetics change is observed in a *Cu-Zn-Al* alloy of an electron concentration $e/a = 1.534$ [4]. The fine strictions of the *18R*-phase appeared first on loading and grew forming a lamellar, multi-interface structure. Then, on further loading a second single-interface *2H*-martensit appears but on an other Schmid factor and grew absorbing this lamellar structure.

The ratio of critical stresses is also 1:2 but these are two times smaller than one showed in Fig. 3. It means, that entropy change of both transformations is also two times smaller, because deformation change was approximately the same.

The kinetics of the multi-interface and single-interface transformations distinguishes through the nucleation and growth. The $\beta \rightarrow 18R$ transformation involves in principle nucleation of more and more martensitic crystals without their growth, whereas the transformation with a single interface after the nucleation represents a pure growth after the nucleation of a martensitic band with a critical

size. The first case at that a number of phase boundaries determines the transformation hysteresis width is described in the Müller's theory [6].

In the second case, where a martensitic band as a critical nucleus is much bigger size, is needed a greater critical shear stress to the transformation beginning. The band growth is a local, not steady process with an own velocity as a internal transformation variable. Transformation locality means that the transformation properties - hysteresis, stress-strain behaviour, depend on properties the specimen at this place, where the single interface is located to this time and depended not on previously transformation history.

The difference is, that the transformation hysteresis by partial cycling is the same as one by complete cycling. The stress-strain behaviour can be more complicated, because it reflects the interaction of the interface with obstacles, composition inhomogenities etc. The transformation is similar to the "dominoes effect". The single-interface transformation echives no thermoelastic equilibrium, because the transformation energy change after the surmounting of a critical nucleation barriere is much more than the elastic coherent energy of the interface.

This energy difference is released as a heat flux and as a kinetical energy of the interface that determinates its own velocity. On the other hand the band nucleation has the typical thermoelastic kinetics.

Conclusions

The macroscopical behaviour corresponded to the single-interface transformation and its kinetics have been investigated with the following results:

1. The stress induced single-interface transformation needs for its nucleation a higher shear stress and the formation of $18R$ phase twins.
2. The stress-temperature dependence of both are distinguished in a slope about two times as it is found in other cases of single-interface transformations.
3. The transformation has no thermoelastic character and is determined as a local transformation.

Acknowledgement. This work is supported by the Federal Ministry for Education and Sciences of Germany.

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