Smart Materials Based on a High and Low Temperatures SME-Alloys

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Abstract. The analysis of the nature of low plasticity of composites with strengthening of hard phases are shown that plastic shear constraint in thin interlayer of the binding phase does not provide the small scale of the structural level of deformation. As a result, a material undergoes a brittle fracture. Therefore, one should find another mechanism for relaxation of internal stress concentrations in composites during loading.

The aim of a present paper is to illustrate the concept of how use a shape memory alloys with phase transformation for metal matrix composites. It has been studied shape memory based alloys and deformation and fracture of MMC with phase transformations in matrix.

It have been shown that if plastic deformation is considered as local structural transformation and instability in the material is provided from the very beginning, this will result in a general increase of plasticity due to a decrease in the scale of the structural level of plastic deformation.

Deformation tests of composites near the phase transition temperature show that there are different transformations induced by a highly non-uniform stress state of the binding phase. Under loading the “quasi-amorphous state” is formed in the binding phase after formation of a microcrystalline, highly misoriented structure with the characteristic size of crystallites less than 10 nm. This structure has high plasticity and strong hardening. It governs an efficient transfer of external load to solid particles, inducing dislocation glide even in typically brittle titanium carbide particles during non-uniform loading.

1. Introduction
A considerable interest has been shown in the last decade in developing ceramic-based materials with high mechanical properties [1, 2]. However, the rate of scientific advance requires a non-traditional approach to developing materials of this type [3, 4]. In most cases, material science was developed as a science of describing the properties of existing materials - either present in nature or developed on the basis of the intuition of investigators. Science only described the properties, tried to find a detailed relationship between the macro- and microscopic properties and, in certain cases, managed to predict them. Attempts to design materials have already been made several years ago when investigations were carried out to develop materials with specified properties [5]. However, important results could not be achieved whilst the methodology of material design remained within the limits of conventional considerations.

The aim of a present paper is to illustrate the concept of how use a shape memory alloys with phase transformation for metal matrix composites. It has been studied shape memory based alloys and deformation and fracture of MMC with phase transformations in matrix.
2. Materials and Experimental Procedure
Materials are prepared by standard methods of powder metallurgy – pressing of powders and sintering in vacuum furnace [6, 7]. It was carried out metallography, scanning electron microscopy of section surfaces, transmission electron microscopy and X-ray analysis. Mechanical tests were carried out using Instron-1185 unit with strain rate under loading $5 \times 10^{-3}$ sec$^{-1}$.

3. Results and Discussion
As indicated by theoretical examination, it may be expected that in a crystal deformed in the vicinity of the point of a structural phase transition a fragmented structure will appear as a result of formation of rotations of the microregions from the very start of deformation. In fact, if layers of a substance with a structural phase transition are placed between the undeformed carbide particles in a composite of the hard alloy type, deformation of this composite material will be characterized not only by the, uniform compression test state of the interlayer by also by its shear deformation as a result of rotation of the individual carbide grains.

As well-known [8] TiNi alloys near equiatomic concentration has a different states, in Table 1 are presented an initial state and temperatures of phase transformations for different TiNi alloys. On Fig. 1 are shown “stress-strain” curves of studied alloys and one can see that it is possible to obtain a very different its mechanical properties.

Table 1. Phase contents and transition temperatures of TiNi alloys.

| Alloy | $M_s$, K | $T_R$, K | Initial state (at room temperature) |
|-------|---------|----------|-----------------------------------|
| 1     | 318     | ≤310     | B19'+B2(R)                        |
| 2     | 295     | 302      | R+B19'                            |
| 3     | 293     | ≤290     | B2(R)+B19'                        |
| 4     | 263     | 283      | B2                                |

On Figure 2 are shown X-ray pattern for alloy 1 (a) and 4 (b) obtained in situ during deformation. As one can see the initial phase content of alloy will stipulate a way of transformations and we can obtain direct B2=>B19 or B2=>R=>B19 phase transform [9].

On Figure 1 are shown “stress-strain” dependencies for near equiatomic TiNi alloys.
Figure 2. Synchrotron X-ray Pattern ($\lambda=0.14879$ nm) for alloy 1 (a) and 4 (b) obtained in situ during deformation.

The experiments with composites were carried out with a sintered composite, in which the role of hard undeformed particles was played by titanium carbide particles 5-10 mm in size, situated in an intermetallic compound TiNi characterised by a thermoelastic martensitic transformation. The system was deformed in the vicinity of the critical temperature of the martensitic transformation and at temperatures considerably higher than this temperature. In cases in which the binding phase was stable, its deformation took place by dislocation slip on main shear planes, mainly (110). The yield limit was inversely proportional to the distance between the carbides and in thin layers of the binder in alloys with a higher content of the hard phase the yield limit was not reached up to failure or was too high to enable sufficient stress relaxation to take place. Beyond yield the limit of these composite cracks appeared at the carbide-carbide and carbide-binder boundary and failure was catastrophically brittle. So, in TiC-TiNi composites we will have different phase states of binder in composite which will stipulate a non-uniform state of TiNi.

The limiting plastic strain of the composites with TiNi in the stable state was low, less than 2%. The limiting strain to failure of alloys with TiNi in the transformation region was 2-3 times higher than that of the alloys with the binder in the stable state, with the strength unchanged.

Deformation of the TiNi system during the loss of shear stability by the lattice is accompanied by large changes of the structural state. Already in the initial undeformed specimens of the TiC-TiNi composite the structure of binder was highly heterogeneous. The presence of a characteristic contrast in the form of ripple marks on bright field electron microscopic images indicates that the TiNi is in the pre-transition state.

Figure 3. (a) - The electron diffraction of binder in deformed composite in initial stage of deformation. (b) - The quasi-amorphous state of a binder before fracture. (c) - TEM picture of TiC particles after composite deformation. One can see the dislocations.

During loading the examined composites transformations were detected in TiNi. The nature of these transformations greatly differed as a result of a highly heterogeneous stress state of the binder in the composite of this type. The relationships governing the variation of the structure may be described as
follows. In loading in the region of elastic behavior of the composite the microstructure of TiNi varied from the dispersed domain structure to a structure with a banded contrast typical of intermediate shear structures.

Electron diffraction patterns show diffusion scattering and then extra reflections in both commensurate and incommensurate positions with a different incommensurability parameter in different directions of the reciprocal lattice. This indicates that several variants of premartensitic domains with a different domain structure in every domain form. This type of transformation in TiNi is caused by a highly complex stress state formed around solid particles during loading the composite. Under the conditions of steep stress gradients formed in the matrix, the directions of displacement of the atoms in microregions, causing a local loss of stability of the B2-structure, are determined by the stress conditions existing at the given moment of loading in the given microvolume of the binder [10].

These conditions also determine the orientation of the newly formed martensitic domains. This nature of transformation in titanium nickelide results in a simultaneous reduction of the peak and integral intensity of the line of the B2 phase on x-ray diffraction patterns, but is not accompanied by the growth or appearance of new martensitic peaks. Only widening of the most intensive lines of the monoclinic nickel-titanium to single-diffusion reflection is observed. This is typical of formation of a fine dispersed deformation structure, consisting of misoriented fragments of the B2 phase and martensitic domains. As shown by the transmission electron microscopy results, the degree of misorientation of these fragments and domains exponentially increases with increasing strain. This is also accompanied by rotation of individual grains of the carbides, and the extent of this rotation increases with a reduction of the size of these grains.

Thus, a fine crystalline greatly misoriented structure with a typical crystallite size smaller than 100 A forms during deformation in the binding phase of the TiC-TiNi solid alloy as a result of the B2 => B2+B19' inhomogeneous transition. This structure is characterised by high toughness and hardening capacity and enables an effective transfer of the external load to the hardening agent, causing in this case dislocation slip, even in typically brittle particles of titanium carbides. A subsequent increase of the load results in multiple cracking of plastically deformed solid particles of TiC with the integrity of the material unaffected. In the final analysis, the fracture toughness is high.

It is well-known that high-strength materials are characterised by brittle fracture. There a large number of fracture criteria of the material used to evaluate the strength. Polynomial criteria of strength in stresses, which formally coincide with the Mises-Hill-type plasticity criterion, are used in most cases. Taking this into account, to solve the given problem, we used the model of an elastic ideally plastic solid with the estimate of the onset of plasticity in the first approximation (start of failure) according to Mises criterion.

Under loading, such a matrix undergoes a martensitic-type phase transition which starts in regions of stress raisers and leads to extensive shape changes in these local regions. Shape changes of this type are, in fact, inelastic deformation which causes effective relaxation of high local stresses. On the other hand, the tough behavior of the matrix supports the transfer of load to hard inclusions. Thus, the mesovolume of the material forms a damping structure capable of dissipating the supplied energy and withstanding high loads as a result of a reduction of the level of stress concentration and redistribution of stresses inside the mesovolume. For comparison of mechanical behavior studies were carried out for a material with a damping matrix, capable of phase transformation (Fig.4, curve 2), and with a conventional matrix (curve 1) in which brittle fracture starts after a limiting strain.

![Figure 4. Stress-strain curves for TiC-TiNi composite](image-url)
Figure 4 shows the pattern of fracture of mesovolumes of the material with different matrices for two consecutive times. Multiaxial loading was applied (combination of tensile, compression and shear loading).

If for the case with a conventional matrix the material may be regarded as completely fractured, then for a structurally unstable matrix, regardless of fractures of inclusions, the material does not lose completely the load-carrying capacity because the matrix is only slightly damaged. The material of the matrix appears to flow around the upper large grain and the extent of fracture of this grain in subsequent stages remains low.

The main work of deformation for the case of the structurally unstable matrix is concentrated in the matrix material. The work of deformation shown in Fig. 3, clearly indicates the characteristic structural elements of the mesovolume and its high value in the central grain is linked with the work of pre-fracture and development of shear deformation in the grain (see Fig.3c).

A very interesting pattern of strain localization (and, consequently, subsequent fracture) due to the heterogeneity of the structure (Fig. 5a) formed in the mesovolume of the ceramic composite during its tensile loading. In the initial stages of deformation in which all stresses are relatively low and elastic, this localization process is not yet completely visible. With increased loading, the strain becomes localized in the direction of the two main tangential stresses.

![Figure 5](image)

**Figure 5.** (a) - TEM image of composite after loading; (b) - The rotation of a carbides vs. its sizes; (c) - The angles between the fragments of binder after deformation

The final stage - design of the material - is the solution of an inverse problem: selection of the optimum physico-mechanical characteristics of the materials of the matrix and solid inclusions and their link with the matrix material, determination of the optimum concentration of hardening particles, etc. However, these applied problems can be solved only by taking into account; the specific features of application of the ceramic composite: the materials are to be used for producing dies, cutting tools, or for other purposes.

4. Conclusion

Thus, on the basis of the general concept of the material as a hierarchically organized system of structural levels of strains and fracture we have developed a model of a structural ceramic composite. The representative mesovolume of this material, being a macroparticle (after averaging the parameters over the mesovolume) is in satisfactory agreement with macro-experiments. On the other hand, the special features of deformation and fracture of the mesostructure can be examined in detail by calculations. For example, calculations should show the development of mesovortex structures and strain localization in the meso-volume. The proposed method may be used directly in designing ceramic materials for specific components.

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