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Large strain burst induced by martensitic transformation in austenitic micropillars

O.G. Nimaga [a,b], G.J. Cheng [c], H.W. Yen [c], M.X. Huang [a,b,*]

[a] Shenzhen Institute of Research and Innovation, The University of Hong Kong, Shenzhen, China
[b] Department of Mechanical Engineering, The University of Hong Kong, Pokfulam Road, Hong Kong, China
[c] Department of Materials Science and Engineering, National Taiwan University, Taipei, Taiwan

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ABSTRACT

The present work investigates, for the first time, the effects of martensitic transformation as well as orientation on the stress-strain behaviour of austenitic micropillars. It is found that large strain burst can be induced by martensitic transformation, and its magnitude depends on the orientation of pillars. Pillars with an orientation close to [1 0 0] show a much larger strain burst than the ones with an orientation close to [1 1 0]. This is because [1 0 0] pillars have a much higher fraction of transformed martensite to which the magnitude of strain burst is directly correlated.

The demands on understanding mechanical properties of micron-sized components in MEMS/NEMS industry are one of the driving forces to investigate the micromechanics in single crystal micropillars. Deformation behaviours of micropillars are being actively researched in many different materials [1–9]. There are generally three main plasticity mechanisms for metals and alloys, namely dislocation plasticity, twinning plasticity and martensitic transformation. Dislocation plasticity in micropillars are generally reported to show stochastic strain burst and size effect [1]. In general, three mechanisms are used to explain the strain burst phenomenon and the size effect related to dislocation-dominated plasticity, namely the dislocation starvation model [10], dislocation source model [11], and the statistical mechanics-based theory [12]. In addition to dislocation plasticity, twinning-dominated plasticity in micropillars also possesses strain burst and size effect [13–16]. For martensitic transformation in micropillars, majority of the existing studies are limited to shape memory alloys (SMAs) [17–21] while only few studies [22] investigated martensitic transformation in austenitic steel. Martensitic transformation in SMAs concerns the reversible super-elastic behaviour that is fundamentally different from the irreversible martensitic transformation in austenitic steels. i.e. face-centred cubic (FCC) austenite transforms to body-centred cubic (BCC) or body-centred tetragonal (BCT) martensite.

Martensitic transformation in austenitic steel is found to preferably form in certain orientations [23–25]. Numerous reports have shown that during deformation martensitic nuclei are formed within the intersection of partial slip bands [26]. The dependence of martensitic transformation on grain orientation is similar to that of the formation of mechanical twins or Shockley partial dislocations [27]. Moreover, it has also been reported that martensitic transformation is not found in certain grain orientations. This is attributed to the infeasibility to form partial slip bands to act as nuclei formation sites [28]. In other words, martensitic transformation is similar to dislocation plasticity and deformation twinning with which all of them are related to dislocation activities and thus depends on crystal orientations. Such similarity between these three plasticity modes suggests that martensitic transformation in micropillars may also exhibit orientation-dependent strain bursts similar to dislocation and twinning plasticity, which has not yet been investigated in literature. The aim of the present work is therefore to study the strain bursts induced by martensitic transformation in austenitic micropillars. The effect of orientations on the strain burst will also be investigated.

The chemical composition of the steel investigated is Fe-0.6C-9.5Mn (wt%). A sample with a dimension of $10 \times 10 \times 1$ mm was sliced from a bulk sample that was solid solution treated at 1273 K for 1 h and quenched in water. The sample was then polished mechanically down to 1 μm. Electropolishing was then carried out under 12 V at room temperature in an ethanol solution with 25% perchloric acid. Scanning electron microscopy (SEM) and electron backscattered diffraction (EBSD) were performed using LEO 1530 with HKL Channel 5 software. EBSD result shows a single austenite phase in the present steel. Grains with orientations close to [1 0 0] and [1 1 0] normal to the sample surface were

* Corresponding author.
E-mail address: mxhuang@hku.hk (M.X. Huang).

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Table 1
Schmid Factor of the primary and secondary slip plane for the two chosen pillar orientations.

	Full dislocation			Partial dislocation			
				Leading partial		Trailing partial	
	Slip plane	Slip direction	Schmid factor	Slip direction	Schmid factor	Slip direction	Schmid factor
[1 1 15] Representing [1 0 0]	(111) $(1\bar{1}1)$	$\langle\bar{1}01\rangle$ $\langle 101\rangle$	0.44 0.44	$\langle 11\bar{2}\rangle$ $\langle\bar{1}1\bar{2}\rangle$	0.5 0.47	$\langle\bar{2}11\rangle$ $\langle 211\rangle$	0.26 0.30
[4 1 6] Representing [1 1 0]	(111) $(\bar{1}11)$	$\langle\bar{1}01\rangle$ $\langle 101\rangle$	0.44 0.46	$\langle 11\bar{2}\rangle$ $\langle\bar{1}1\bar{2}\rangle$	0.30 0.31	$\langle\bar{2}11\rangle$ $\langle 211\rangle$	0.45 0.48

selected for fabrication of micropillars as these two orientations favour partial dislocation slip and full dislocation slip, respectively. In the following, [1 0 0] and [1 1 0] will be used to represent the actual orientations of the pillars. The actual orientation and corresponding Schmid factors for the primary and secondary slip systems are provided in Table 1. The stacking fault energy (SFE) of the current steel is calculated, using ThermoCalc, to be roughly 12.4 mJm^{-2}, assuming the interfacial energy of the stacking fault to be 5 mJ m^{-2} [29]. For the [1 0 0] orientation, the Schmid factor of the leading partial is larger than that of the trailing partial so that the glide of partial dislocations is most likely to occur in the [1 0 0] pillar. Whereas, for the [1 1 0] orientation, the Schmid factor for the trailing dislocation is larger than that of leading partial so that full dislocation slip will most likely occur [7]. Micropillars were fabricated by focused-ion beam (FIB) using FEI Quanta 200 3D. During the milling process, a current of 20 nA is used in the initial step to fabricate a trench of 50 μm, followed by subsequently smaller current, down to a value of 0.5 nA. Pillars with a large diameter of 8 μm are employed in the present study in order to reduce the extent of stochastic nature of small pillars [15]. The height-to-diameter ratio of the pillar is maintained between the values 2–3, with tapering angle kept within 5°. For each orientation, 8 pillars were prepared and compressed by a 20 μm flat punch indenter using Nano Indenter G200 at a constant loading rate in the load-controlled mode. The loading rate of 0.24 Nm/s is equivalent to the strain rate roughly of 10^{-4} s^{-1} and all samples were compressed to the same final load. After compression, transmission electron microscopy (TEM) samples were cut out longitudinally along the pillars using FIB. TEM investigations were carried out using FEI Tecnai G2 F20 TEM at an operation voltage of 200 kV.

The engineering stress-strain curves of pillars with two different orientations are shown in Fig. 1(a). In total, there are 8 samples for each orientations involved in the analysis. It is noted that only 4 stress-strain curves are shown for each orientation for illustration purposes. Similar to many reported micropillar experiments, stress-strain curves initially show a linear stress-strain relation until the stress reaches a critical value where the first strain burst occurs [4]. For the [1 0 0] pillars, it shows a lower stress for the first strain burst compared to that of [1 1

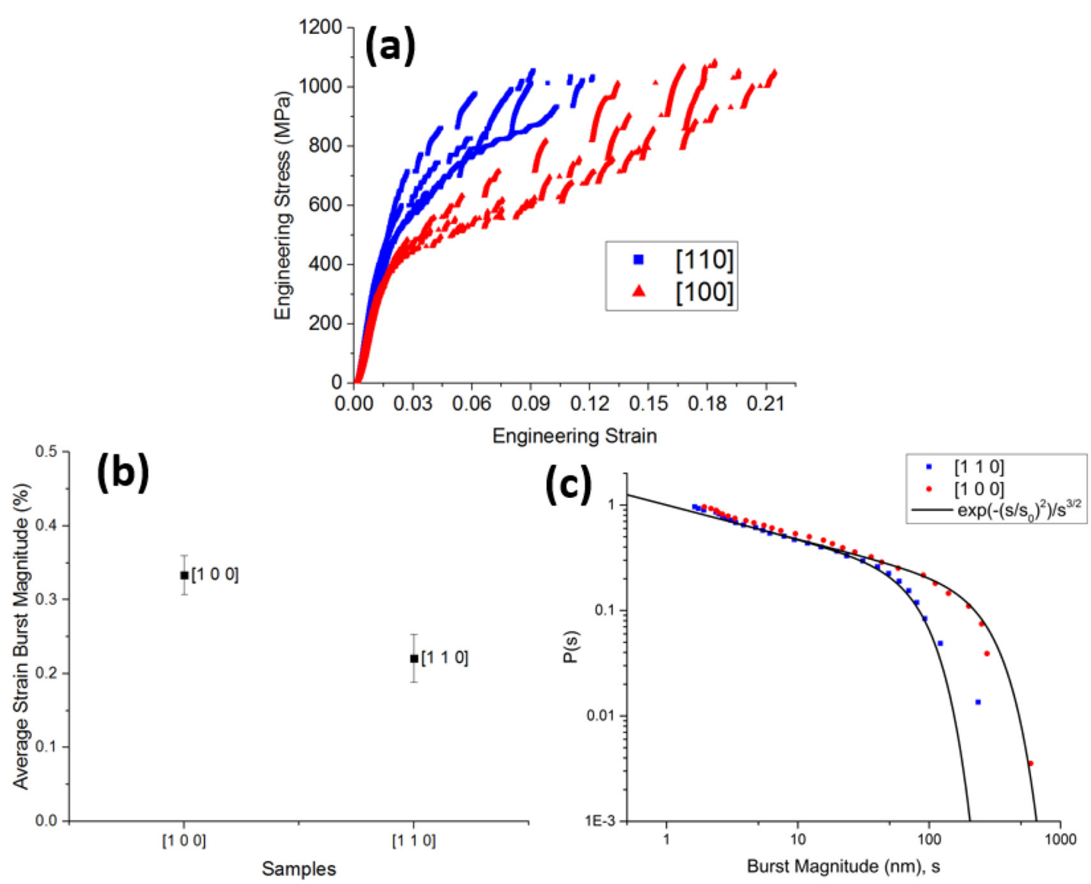

Fig. 1. (a) The engineering stress-strain curves of [1 0 0] and [1 1 0] micropillars; (b) The average strain burst magnitude of [1 0 0] and [1 1 0] micropillars; (c) The statistical distribution of burst magnitude of [1 0 0] and [1 1 0] micropillars.