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Origin of the Bauschinger effect in a polycrystalline material

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ABSTRACT

There is a long and lively debate in the literature about the origin of the Bauschinger effect in polycrystalline materials, the most widely accepted explanation being the easier movement of dislocations during reverse loading causing the reduction of the yield stress. Other explanations include incompatible deformation at the grain scale and change of dislocation cell structures during forward and reverse loading, but recent publications show these phenomenological explanations of the Bauschinger effect are not holistic. In the experimental work presented here, we have investigated the role of micro residual lattice strain on the origin of the Bauschinger effect in type 316H austenitic stainless steel using in-situ neutron diffraction. Standard cylindrical specimens were tension-compression load cycled at room temperature with the loading interrupted at incrementally larger compressive and tensile strains followed by reloading to the tensile loop peak strain. Mirror symmetric cyclic tests were also performed with tensile and compressive load interruptions followed by compressive reloading to the compressive loop peak strain. A strong correlation is demonstrated between the evolution of residual lattice strain in the grain families and the change in magnitude in macroscopic yield stress, peak stress and the shape of the yielding part of the stress-strain curve for both the cyclic tension yield and compression yield tests. This implies that the residual lattice strain generated by grain scale elastic and plastic deformation anisotropy is the primary source of the Bauschinger kinematic hardening effect observed in type 316H austenitic stainless steel.

1. Introduction

Kinematic hardening of steel during tension-compression cyclic loading was first described by Bauschinger [1] in 1886. In his paper he established the following two laws [2]:

- If a bar is loaded beyond its elastic limit, this increases the elastic limit of the bar for a subsequent load in the same direction.
- Plastically deforming a bar in tension or compression reduces the elastic limit of the bar for a subsequent compression or tension loading respectively. The more the bar has deformed plastically the greater the reduction in the reverse loading elastic limit.

The Bauschinger effect can significantly affect the deformation response of materials and structures which experience tension-compression cyclic loading, by altering the material's elastic limit during the reverse cycle. Such effects can, therefore, be important for life estimation of metallic components in practical applications, for example, power generation plant components. Inelastic constitutive models (for example [3]), are usually used to predict the evolution of stress-strain during such cyclic loading. Such models often use Bauschinger tests as means for characterising work hardening in the material. Therefore in order to develop an accurate inelastic constitutive model for a material it is critical to understand the mechanisms controlling the Bauschinger effect.

The Bauschinger effect has been observed in many single crystal and polycrystalline metallic materials including aluminium [4], nickel [5], steel [6,7], copper and brass [8]. Many investigations have been conducted exploring the origin of the observed Bauschinger effect in these and other metals. However, the number of models or theories that have been developed to explain the Bauschinger effect is almost equal to the number of investigations [9]. The theories can be divided into two overarching groups; continuum and microstructure based. Generally speaking, continuum theories describe the Bauschinger effect in terms of back stresses, generated due to inhomogeneous deformation in the material, while microstructure based theories describe it in terms of dislocation structures and those interactions with other dislocations, precipitates etc. in the material's microstructure during forward and reverse loading.

The most accepted microstructure based explanation for the Bauschinger effect of Orowan [10] and Sleeswyk [11] describes the mechanism in terms of easier reversibility of dislocation motion upon

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load reversal. Among notable theories, Mughrabi [12] proposed that incompatibility of deformation between harder cell walls and softer cell interiors of dislocation cell type structures caused the Bauschinger effect. It is well evidenced that formation of dislocation cell type structures induces internal stress in the material [13–15], and a number of investigations [12,16–18] have been conducted correlating such stresses with the Bauschinger effect. Moreover, a number of semi-phenomenological theories [16,18–26] have described the correlation between the changes in dislocation structures during reverse loading and the material's hardening behaviour. These models postulate formation of dislocation structures and dissolution of the structures during reverse loading as the primary cause of the Bauschinger effect in metals. However, such theories of dislocation cell type structures in polycrystalline materials have been contradicted by others, for example [20,27,28,29,30,31]. In the light of these recent publications, it can be stated that dislocation cell type structures and changes of these structures during the cyclic loading direction change have no unique relationship with the observed Bauschinger effect.

Recently, the contribution of residual lattice stress (also termed type II residual stress, intergranular stress, internal stress, micro residual stress and misfit stress) to the Bauschinger effect has been discussed [32,33,34]. Lattice residual stress/strain originates due to a combination of elastic and plastic anisotropy of deformation at the grain scale. Its potential role in the kinematic hardening of metals was predicted a long time ago [35,36]. However, there is little published experimental evidence characterising the relationship between micro residual stress and observed macroscopic deformation in polycrystalline materials.

In this paper, we present results from a set of uniaxial cyclic stress-strain tests which investigate the relationship between the evolution of internal lattice residual strains and observed macroscopic Bauschinger effect in AISI Type 316H austenitic stainless steel at room temperature. In-situ neutron diffraction is used to measure the evolution of residual strains in variously oriented crystallographic grain families after being interrupted at various points in the forward loading cycle and correlating these residual strains with the observed change in macroscopic yield stress during the following reverse and forward loading cycle. We believe this is the first time that a direct correlation between micro residual strain and the Bauschinger effect has been quantified experimentally.

2. Material and experimental technique

The cyclic stress-strain specimens for the experiment were machined from as-received AISI type 316H austenitic stainless steel bar and then solution heat-treated (1050 °C heat soak for 1 h followed by a water quench). The cylindrical test specimens had a gauge diameter of 8 mm, a gauge length of 14 mm and a total length of 74 mm.

Lattice strain measurement using neutron diffraction is a well-established technique [37]. For the present study, in-situ neutron diffraction was used to measure the evolution of lattice strains during cyclic deformation. The experiment was conducted using the Engin-X Time of Flight (ToF) neutron diffraction instrument [38] at the Rutherford Appleton Laboratory, UK. The ToF technique allows the change of lattice spacing between crystallographic planes to be measured for variously oriented grain families simultaneously. The measured spacings are then converted to elastic lattice strain values for specific grain families. The Engin-X instrument is equipped with a stress-rig for in-situ experiments which allowed measurement of the evolution of lattice strains as a function of applied load and strain path. A neutron gauge volume of $4 \times 6 \times 4\ mm^3$ was used with an average neutron counting time of ~ 300 s per measurement point. The macroscopic strain was measured using an Instron extensometer placed in the gauge length of the specimen.

3. Experimental plan

The aim of the first test was to investigate the changes in cyclic tension yield stress with respect to various load interruptions in tension and compression, hence this experiment is denoted ‘tension yield test’. The aim of the second test was to investigate the changes in cyclic compression yield stress again with respect to various load interruptions in tension and compression interruptions, accordingly, this second experiment is denoted ‘compression yield test’.

First, a type 316H austenitic stainless steel specimen was subjected to a small stabilising load (5 MPa stress) and the reference lattice spacing (d_0) between lattice planes measured for a number of crystallographic grain families. A grain family consists of a set of grains with similar crystallographic orientation, of which the corresponding *hkl* plane normal lies parallel to the scattering vector of the diffraction. The specimen was then subjected to continuous tension-compression cyclic loading, starting in tension, with a strain rate of $7 \times 10^{-6}\ s^{-1}$ over a total strain range of ± 1% up to the point of saturation of cyclic isotropic work hardening in the material. Upon reaching an isotopically saturated work hardened stage, the loading of the specimen was interrupted at various tensile and compressive strains, such that, a pronounced Bauschinger effect was observed in the material's yielding behaviour during reloading cycles. The interruption strains were chosen systematically at an interval of 0.2% strain. The interruption points of

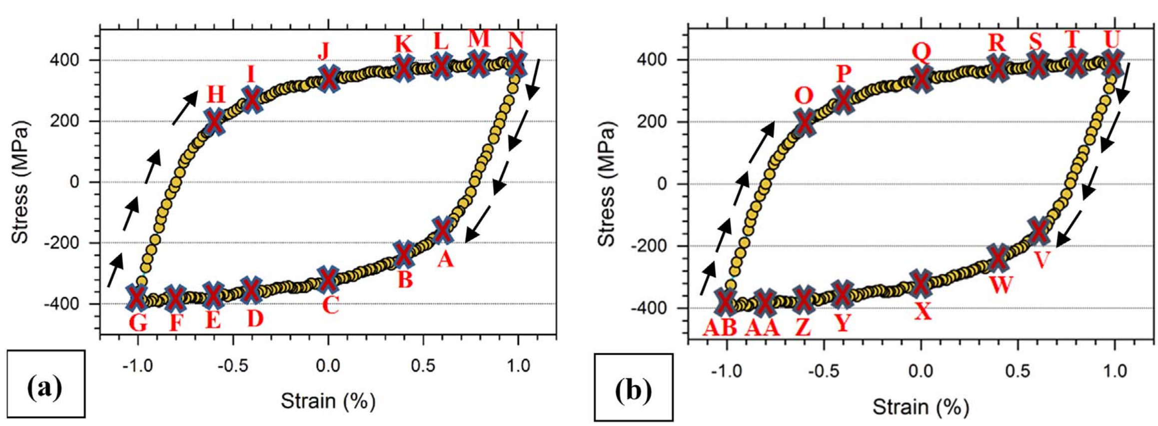

Fig. 1. (a) Interruption points for tension yield test and (b) compression yield test shown in a typical saturated cyclic stress-strain curve of type 316H stainless steel. In (a), points A–G were interrupted while going into compression from a tensile peak strain of 1%. The specimen was reloaded to tensile peak strain after these interruptions. Points H–N were interrupted while going into tension from a compressive peak strain of − 1%. The specimen was unloaded to near zero stress (~ 5 MPa) and reloaded back to tensile peak strain of 1% after these interruptions. In (b), points O–U were interrupted while going into tension from a compressive peak strain of − 1%. The specimen was unloaded to compressive peak strain after these interruptions. Points V–AB were interrupted while going into compression from a tensile peak strain of 1%. The specimen was unloaded to near zero stress (~ 5 MPa) and reloaded to compressive peak strain of − 1% after these interruptions. The black arrows show the direction of loading.