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The Mechanised Testing and Sequential Wear-Analysis of Replica Bronze Age Palstave Blades

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There have been few attempts to conduct highly-controlled laboratory experiments to isolate how wear propagates on metal artefacts with differing metallurgy during simulated use. This reflects a lack of appreciation for the underlying structure of materials within the production of reference datasets for metalwork wear-analysis. Here, we present the use of a drop tower (Instron CEAST 9350) to reconstruct use on replica Bronze Age palstave axes with archaeologically-relevant microstructures. The development, form, and properties of surface wear at the cutting-edge were sequentially analysed by low-power microscopy (digital), high-power microscopy (Scanning Electron Microscope), and microhardness indentation. Major deformations of the blade were documented by photography. This intensive approach reveals the impact of abrasive wear associated with sharpening and use, as well as the frequency and morphology of larger deformations generated by repeated impact – all of which, we demonstrate, can be overtly modified by subtle differences in metallurgy.

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Introduction

The treatment of an object throughout its period of consumption (use) provides vital insight about how its role was embedded within society (Marshall, 2008, pp. 63–65). By studying the patterns of wear made to materials by mechanical processes during use, it is possible to obtain information about the application of individual artefacts, such as, the opposing material, or the nature of contact and use (Bradfield, 2015). Although examination of wear on archaeological material has been routinely applied to lithic objects for over half a century (Semenov 1964), the complexity of metal, a material which can be so incredibly altered by compositional differences, manufacture, reshaping, recycling, and corrosion, has led to a relative reluctance to investigate metal artefacts for evidence of wear (Roberts and Ottaway, 2003, p. 120). Despite these obstacles, a consolidated approach for metalwork wear analysis has since been advocated by Kienlin and Ottaway (1998); this method involves the simulation of task(s), by which wear was introduced on the object in the past, during experimental trials with replicas, and the creation of a repository of marks to identify similar features on corresponding ancient specimens. The application of reconstruction experiments with replicas and/or existing experimental datasets have subsequently become an integral part of metalwork wear studies (Bridgford, 1997, 2000; O'Flaherty, 2002, 2007; Roberts and Ottaway, 2003; Moyler, 2007; Brandherm, 2011; Dolfini, 2011).

Nonetheless, the extent to which the complexities of metal, as suggested above, have been truly accounted for within reconstruction experiments for metalwork wear analysis remains rather limited. Often studies have failed to appreciate the relevance of the microstructure within their experimental scheme – some completed only the characterisation of past materials (Gordon 1985; Bridgford, 2000; Modlinger 2011), or modern replicas (O'Flaherty et al., 2011), while next to none have provided comparisons between the two (Anderson, 2011). Furthermore, with the exception of Kienlin and Ottaway (2008) and Soriano-Llopis and Gutiérrez Sáez (2007), very few studies have attempted to document the progression of wear on replica artefacts with different alloy compositions and manufacturing during experimental work. Sequential experiments are necessary for determining how, or at what rate, wear striations and deformation processes propagate on replicas with a range of archaeologically-relevant microstructures. When combined with high-power microscopy techniques, like Scanning Electron Microscopy, sequential sampling may present a powerful tool for better detailing the mechanisms by which edge-rounding, micro-cracking, and striation marking develop, and how their characteristics relate to type, or duration of use (Ollé and Vergès, 2014; Dolfini and Crellin, 2016, p. 84). To permit the acquisition of comparable results across samples, a highly-controlled experimental setup must be adopted to minimise the impact of variation within the applied force, executed action, and opposition material (Bamforth, 2010, p. 101). It may be necessary, therefore, to undertake experimentation using specifically-adapted mechanised rigs to simulate highly-specialised scenarios (Bridgford 1997, 2000; O'Flaherty et al., 2011).

Hence, the aim of this research project was to establish a multi-method approach to model the progression of surface wear and the appearance of major deformation phenomena on archaeologically relevant tin-bronze microstructures, with the purpose of developing measurements of use-intensity (the amount of use) at a highly detailed level. To this end, the common methodological approach of metalwork-wear analysis has been somewhat altered. Firstly, controlled experiments in a mechanical rig (instead of actualistic field tests) and sequential sampling are adopted here-in, so that the rate at which wear develops, and the features that are linked to material degradation (rather than misuse or random damage), can be accurately recorded. Secondly, we make attempts to record wear at higher levels of magnification, to allow for a better understanding of the way degradation occurs at the micro-scale. When transferred to ancient metals, this data may help to elucidate the use-intensity of these artefacts at a much finer scale of inference, changing the focus from used/not-used, to the length of functional biography before discard or recontextualisation (Joy, 2009, p. 545). Analysis of this sort could, therefore, reveal trends in the narratives of these artefacts and highlight the presence of outliers that may suggest the existence of other unique and diverse customs.

The objects at the centre of the present study are palstave axes originating from Bronze Age Britain and Ireland. Given the emphasis placed on encapsulating the distinctive microstructures of metal objects above, the focus placed on palstave axes from a confined geographical area, as implied here, is an attempt to prioritise the thorough and accurate determination of the specific tribological system pertaining to these artefacts. Emerging to prominence during the period 1500-1000 BC, the palstave axe represents the third major typological development in bronze axe morphology during the Bronze Age in Britain and the first axe type to include a stop-ridge. Previous studies have detailed traces of woodworking on a number of variants of Bronze Age axes: Scottish flat axes (Moyler, 2007); British and Irish flat axes (Crellin, 2017); flanged axes of the North-Alpine region (Kienlin and Ottaway, 1998); and, socketed axes of Yorkshire and Scotland (Roberts and Ottaway, 2003). In each of the aforementioned studies, a considerable proportion of axes were found to demonstrate evidence of use. However, palstave axes have received very little attention from metalwork wear analysts. This is, in part, related to the largely unused condition that palstave axes are often found in (Knight, 2018, p. 430), as well as long-standing ideas about their possible multi-functional purpose (Harding, 1975, p. 517). If anything, such preconceptions work only to raise the potential for unexpected and interesting revelations regarding the use of these objects.

Structure and wear-characteristics of MBA palstave axes

Wear is the culmination of both deformation and chemical processes on the surface of a material when it is in relative motion with the surface of another substance. To understand the nature of wear and, ultimately, how it propagates on bronze objects, the tribological system of the material must first be understood. This requires familiarisation with the structural or mechanical components that dictate the contact conditions between two surfaces in relative motion, which, in-turn, contributes to forming system-related wear characteristics.

In the Middle Bronze Age, a consistent method of production and fine-tuning seems to have been applied to palstave axes by most bronze smiths; this was, smelting, alloying, casting, followed by hammering, localised annealing, a final hammering, and sharpening (Coghlan et al., 1970). Copper was intentionally alloyed with tin, lead, and possibly arsenic, during the course of the Bronze Age in varying quantities and combinations. Tin was the preeminent alloy choice throughout much of the Middle Bronze Age in Britain, when palstaves were the dominant form of axe. Needham et al. (1989 p. 392) have suggested that tin was increasingly adopted for the production of copper objects between c. 2200-2000 BC, and many contained as much as 10-16% tin by the transition to the Middle Bronze Age. Combined data from the analysis of twenty-eight palstave axes from across Britain and Ireland spearheaded by Coghlan (Coghlan and Cook, 1953; Coghlan et al., 1970; Coghlan, 1970a, 1970b) indicates an average tin content of 9.7%. While unpublished compositional data produced by Peter Northover for 112 palstave axes from the south-west of England suggests a mean tin content of 13.07%. Brown and Blin-Stoyle's (1959, p. 190) analysis of 438 pieces of bronze metalwork dating to the Middle Bronze Age and Late Bronze Age revealed an average tin composition of 11%.

The adoption of the bi-valve mould for the casting palstave axes meant that after the sprue and any 'flashes' of metal were removed (by 'fettling'), the as-cast form was often very close to the desired shape of the tool (Rowlands, 1976, p. 71; Kienlin, 2013, p. 427). The use of metallographic techniques has permitted the exploration of the microstructure of archaeological metals, contributing to the understanding of palstave axe design and manufacture. Metallographic analysis suggests that palstave axes were not subjected to extensive forging, though some moderate hammering, in combination with annealing, is generally apparent at the blade (Coghlan and Cook, 1953; Coghlan et al., 1970; Coghlan, 1970a, 1970b). The majority also received a final cold-hammering at the blade edge (ibid.). This manufacturing process has a great impact on the resistance and/or susceptibility of the cutting-edge to macro-scale deformation phenomena like bending, notching, and fracture – this is explained below in further detail.

Alloying of the metal can serve to decrease plasticity and, therefore, increase baseline hardness. Atoms of copper (Cu) are arranged into face-centred cubic lattices. Tin (Sn), on the other hand, forms a tetragonal crystal structure. When tin is added to molten copper, melted, and then allowed to solidify, different phases of crystal may be produced within the microstructure of the alloy (Henderson, 2000, p. 209). Mixtures with low levels of tin predominantly produce a homogenous 'dendritic' alpha phase, where atoms of tin are directly substituted within the face-centred cubic lattice (Wang and Ottaway, 2005). These atoms change the regular distance between copper atoms, generating strain in the structure, corresponding to greater hardness. When the tin content within the copper-alloy is increased, the dendritic crystals are interspersed by darker patches of crystal, the 'αδ eutectoid'. The presence of this phase effectively immobilises many slip planes within the material, so that the copper can no longer be deformed so easily. Thus, the most advantageous quantity of tin within a copper-alloy for producing a material that is of a sufficient hardness to be used to form weaponry or tools, but not so hard that it is liable to brittle fracture, is approximately 10-12%.

Annealing produces recrystallized copper, which has a higher elastic limit and greater fracture toughness than as-cast copper. When copper is subjected to several cycles of cold-hammering and annealing, a final grain size of about 10µm, or even smaller, may be produced. A fine grain size is highly beneficial for increasing fracture resistance as it effectively immobilises dislocations within the crystal lattice, therefore, reducing the potential for plastic flow. This is known as 'Hall-Petch' behaviour (Rubio et al., 2019).

Hammering, as a final treatment, can produce an ever greater hardness in the material by the process of 'work-hardening'. As plastic deformation occurs, and dislocations progress through the crystal lattice, they begin to stack up at grain boundaries, inclusions and other obstructions. During this process the material becomes more difficult to deform, as the number of options for movement are increasingly reduced and, thus, the hardness becomes elevated (Dungworth, 2013, p. 149). Eventually plastic deformation comes to an end when dislocations are unable to travel any further, and the material may fracture if further force is exerted upon it. This process forms the basic premise of many micro-level wear mechanisms, which are generated when the successive loading of a surface "distorts the material around the points of contact", generally leading to the fracture of the material on some level (Mate, 2007, p. 40). The main types of wear mechanism that may be present on the surface of an axe when used in a wood-chopping scenario are likely to be abrasion, adhesion, and fatigue (Bhushan, 2013a, p. 316).

Abrasion occurs during extended periods of frictional contact between solid structures and 'snags' with an equivalent or greater hardness, creating grooves on each surface that correspond with the asperities on the opposing surface (Stachowiak and Batchelor, 2013, p. 525; Bhushan, 2013b, p. 459). The displaced material can become very brittle due to sustained deformation, and may be severed from the main matrix. This type of wear is often expressed on Bronze Age tools in the form of 'striations' at the cutting-edge. Adhesive wear takes place when strong adhesive bonds form between surfaces in relative motion – a particle of the softer material may become detached from its original matrix and transplanted onto the opposition material, creating microwelds (Bhushan, 2000, pp. 278, 279; Takadoum, 2013, p. 83). For example, the well-documented phenomenon of metal adherence to whetstones during the process of sharpening. Fatigue wear, is slightly different, in that it refers to the weakening of the underlying structure of a material. This is often presented as micro-cracks that eventually permeate to the surface (Bhushan, 2000, p. 292), which, in this context, could result in a partial fracture at the cutting-edge of the implement.

Methodology

Six replica axe blades (reconstructions of the cutting-edge only), with differing composition and post-production processing (a summary is presented in Figure 1) were reproduced at Butser Ancient Farm by bronze smith, Jim Clift. The axe used as a pattern for casting the replica blades was a low-flanged palstave with typical features of this type including, a maximum height at the stopridge, a broad-blade that is slightly expanded, a sunken shield motif filled with multiple ribs, and ribs above the stop, on the septum floor (Rowlands, 1976 pp. 32). The decision to experiment with 10 and 14% tin-bronze relates to the existing compositional data for MBA palstave axes, as discussed earlier (Brown and Blin-Stoyle, 1959; Coghlan et al., 1970; Coghlan, 1970a; Coghlan, 1970b; Rowlands, 1976; Needham et al., 1989). It is clear that other elements – low levels of lead and arsenic and values of nickel at approximately 0.4% (ibid.) – are often present within the metallurgy of these objects. However, these additions would only have a minor impact on the material structure and, thus, were not included. A bivalve sand mould was chosen for casting, which was achieved with the use of a modern furnace. Though invisible in the archaeological record, the production of sand moulds has been suggested on several occasions (Ottaway and Seibel, 1998; Ottaway, 2001; Heeb and Ottaway, 2014). Equally, stone, sand, and clay moulds appear to produce comparable dendritic arm spacing and hardness – though the same cannot be said for bronze moulds (Wang and Ottaway 2005, pp. 49, 55).

After casting, the axe was removed from the mould and allowed to cool down without quenching. Cold-hammering consisted of lightly hitting the blade with a steel ball-point hammer. Blades that were subjected to annealing were heated in a furnace for 20 minutes at 650ºC and cooled at ambient temperature. The final sharpening, as well as subsequent re-sharpening events, were administered by the of use of a limestone grinding stone in a circular motion until a reasonably acute angle was formed at the edge. This was followed by a finishing treatment implemented by a sandstone whetstone in a circular motion. While all reasonable efforts were made to produce the experimental specimens in a way that was representative of past methods, there is little doubt that this was not completed with the level of skill held by a prehistoric bronze smith.

The degree of porosity observed varied slightly across the experimental specimens. Similar to most Bronze Age palstave axes, the microstructures of Blade 1, Blade 2, and Blade 6 all exhibited only a small number of pores of a negligible size (Coghlan et al., 1970; Coghlan, 1970a; Coghlan, 1970b). The samples taken from the cutting-edge of Blade 3 were fairly absent of small pores, but one large cavity was discovered in the centre of the blade tip. Blade 4 and Blade 5 demonstrated a network of small or moderately-sized pores. The higher porosity of Blade 4 was expected since there was no cold-hammering administered to 'close-up' the pores. In contrast, Blade 5 was a poorer quality of cast for reasons unbeknownst to the authors. Each of the respective microstructures of Blades 1, 2, 3, 5 and 6 demonstrated recrystallization close to the exterior of the sample (Figure 2). The crystal grains within each sample contained annealing twins and many exhibited strain lines. Across all specimens, the bulk of the material remained in a dendritic form. It appears that the temperature and/or time of annealing applied to each axe was not sufficient enough to generate complete recrystallization of the entire microstructure. Metallographic analyses of the cutting-edge of Bronze Age palstave axes more commonly demonstrate pervasive recrystallization but have also suggested the occurrence of partially recrystallization microstructures (Coghlan and Cook, 1953; Coghlan et al. 1970; Coghlan, 1970a, 1970b). It should be noted, however, that the mechanical advantage offered by partial recrystallization will be somewhat less than full recrystallization.

The experimental rig comprised of an Instron CEAST 9350 impact tower, based at the University of Southampton. This instrument operates by dropping the striker and tup (specimen) onto the opposition material situated in the impact chamber below. A screw thread was inserted into the block part of each Blade to facilitate its attachment to the instrument striker. The impact tower permitted the energy at which the specimen was dropped onto the opposition material to be specified. Based on experiments that involved filming an axe-swing with a high-speed camera, the average energy at impact was estimated to be 15J (Andrews, forthcoming, pp. 117-120). Despite this, after 2000 impact tests with the aforementioned energy value, testing was conducted with higher energy parameters (30J and 45J of impact energy) with Blade 1, 2, and 6. While the executed action produced by the impact tower – a downwards drop onto a flat wooden block – produced similar compressive stress onto the axe-head as in authentic use, it did not reflect the same motion or angle of impact generated during an human arm-swing. This may have the undesired effect of re-directing stress from the blade corners (the area that makes contact first in an axe-swing), to the centre, introducing fewer failures in the aforementioned areas and, thus, potentially leading to an underrepresentation in the progression of asymmetry within the experimental samples.

The opposition material chosen for the study was oak, which was selected for a number of reasons. These include, an observed decline in the number of oak trees in Britain during the Bronze Age period (Birks, Deacon and Peglar, 1975, p. 88; Beckett, 1979, p. 842), the evidence for the use of oak as a structural feature within Bronze Age infrastructure (Drewett, 1979; Main et al., 1998), and the desire to test the replica axes against a relatively hard wood species, so that the maximum amount of wear could be induced and recorded. The experimental blades were used to cut the oak blocks across the grain. While the material is less resistant to compressive forces in this orientation, it is not predisposed to fracturing. This avoided the splitting of the wood, which would endanger the sample and increase the number of oak blocks that would need to be used in experimentation considerably.

The present study has utilised a range of techniques to examine the development, form, and properties of surface wear on replica palstave axes. Surface microhardness indentation was used to investigate the changing mechanical properties of the Blades 1 and 2 as they were progressively used. It was thought that sequential hardness mapping might eventually be used to produce a mathematical model, that upon comparing the corresponding data from prehistoric palstave axes, would permit a quantitative prediction of the number of impacts. Using a load of 500 gf, five measurements were taken at three sampling locations at the cutting-edge, while five further 'baseline' measurements were recorded 1cm from the cutting-edge as a control. Measurements were made before and after sharpening and at intervals of 250 impact tests. Although efforts were taken to make sure the surface was parallel to the indenter, the surface of the axe is inherently curved and many measurements exhibited some form of vertical distortion. Some measurements also exhibited some degree of horizontal distortion due to the largely natural state of the material surface, which was quite rough. These errors will likely have introduced some level of inaccuracy into the calculation of surface hardness.

Scanning Electron Microscopy (referred to as 'SEM' from this point onwards) was adopted to track surface features such as wear marks, edge-reduction, and cracking on Blades 1 and 2. SEM imaging was performed using a JEOL JSM-6500F and JSM6500 software. The vacuum pressure used throughout all imaging was <9x 10-4 Torr, the voltage was 15kV, and the probe current was set to 11. The blades were cleaned with water and soap, then washed in ethanol, and dried, before analysis was undertaken. An ultrasonic bath was not used, since it was an aim of the study to observe the occurrence of any adhesive wear mechanisms taking place, and this sort of treatment would remove most particles stuck to the surface. Analysis was conducted at three pre-determined points of interest at the blade edge (the apex of the blade and two points 1cm either side of this). Images were taken at magnifications of x25, x50, x100, x250, x500, and x1000, before and after sharpening and every 250 impact tests.

Low-power (digital) microscopy was employed in order to reveal the rate at which microscopic striations appear at the cutting-edge on Blades 3, 4, 5 and 6. While digital microscopy does not possess the resolution and depth of focus of SEM imaging, the setup for this technique is portable and does not have a fixed base, which means that it can easily be transported to in-situ locations and large objects can fit beneath the microscope easily. Also, the samples do not need a great deal of cleaning to be imaged, and any organic residues left on the surface do not obscure the features greatly. A Dino Lite 2.0 microscope was used in conjunction with the associated Dino Capture 2.0 software. The cleaning method stipulated for SEM imaging was also used before digital microscopy. Analysis was again conducted at the three pre-determined specified in the above text, before and after sharpening, and every 250 impact tests. Images were taken at x25 and x50 magnification using the 'edge of field' feature to create a range of focal layers that were stitched together automatically by the software. Photography was also employed across all specimens to record the frequency and form of blade failure at the cutting-edge. Photographs were taken using a Fujifilm FinePix F20LE camera.

Results

Preliminary experiments conducted by the authors demonstrated that the initial sharpening induces a great elevation in average hardness (Andrews, forthcoming p. 157). In-line with these results, the mean hardness increase at the cutting-edge of Blade 1 between the last annealing treatment and initial sharpening was 71 HV. The average hardness across the control zone on Blade 1 was also raised considerably (47 HV) as most of the indents were placed on areas that had been affected by sharpening. The results presented for Blade 2 are in a slightly different configuration, as this specimen received a final cold-hammering before sharpening. Thus, the mean hardness increase at the cutting-edge of Blade 2 after final cold-hammering was 28 HV, with an additional increase of 46 HV after sharpening – equating to a total increase of 74 HV. The average hardness across the baseline zone on Blade 2 was raised by 7 HV after the hammer-hardening, and another 60 HV by sharpening.

The surface hardness of Blades 1 and 2 was successfully recorded over 2000 impact tests. The average hardness values recorded at the control position on Blade 1 fluctuated between 135 HV and 150 HV, approximately, while they were in the range of 150 HV and 166 HV for

Blade 2 (this is most definitely related to the final cold-hammering that this specimen was subjected to). As seen in Figure 3**,** the mean hardness recorded along the cutting-edge displays an increasing trend for both Blades 1 and 2. The data shows that a particularly rapid hardening effect was experienced by Blade 1 at the beginning of testing; by only 750 impact tests the sample locations across the cutting-edge exhibited a mean hardness of 200±24 HV and fluctuated around this value for the remainder of the testing. The surface hardening encountered by Blade 2 was more gradual – at 500 impact tests the mean hardness was already at 183±24 HV, but it was not until 1250 impact tests when the measured values were close to 200 HV when averaged.

It is possible to say with some confidence, given the results presented above, that the surface of the cutting-edge undergoes a hardening effect during use. This appeared to be very encouraging in terms of establishing data for the surface hardness measurements recorded on ancient specimens to be compared against. A Bronze Age palstave axe (an unprovenanced 'Side-Flanged' palstave found by a metal detectorist) was used to trial the application of surface hardness testing to a prehistoric bronze axe (Andrews, forthcoming, p. 164). Unfortunately, upon loading, the indenter broke through the oxidised surface of the axe and onto a corrosive layer beneath, in which the indents were completely engulfed. Not only was it not possible to measure the indent, but the material matrix had clearly been completely compromised and any values that could be extracted would not reflect the original hardness of the surface. Thus, considering that this palstave was a very well-preserved specimen, it seems unlikely that surface hardness testing as a proxy for use-intensity is a viable option. Nonetheless, this does not render this data useless; quite the contrary, the results provide useful quantitative evidence for the likely embrittlement of the cutting-edge, the effect of which has been documented visually in the SEM imaging, discussed below.

The SEM data has permitted many observations to be made regarding the transformation of the surface of the blade during use, and the specific microscopic wear processes at hand. To begin with, the cutting-edge sustained several changes in profile throughout the experimental process. Prior to sharpening, the edge contained smooth irregularities which were introduced due to the imprecise nature of the casting process. After sharpening, the cutting-edge was quite significantly reduced, and appeared to be well levelled at lower levels of magnification. However, upon greater magnification, it is clear that the cutting-edge was altered towards a much rougher condition. As to be expected from the area sustaining the most impact force, the edge presented clear modifications from an early stage of impact testing. Sequential imaging at x100 magnification documented the rapid deterioration of the cutting-edge in some areas, exemplified by the formation of large cavities at the very tip of the blade by 750 impacts (Figure 4c). At the same time, any prominent edge features experienced significant flattening by 1000 impact tests.

As demonstrated by Figure 4a, the surface of both Blades 1 and 2 was left in a mottled state after casting. It is evident that sharpening has produced clear horizontal lines (mostly) parallel to the blade edge (Figure 4b). At magnifications of x25, x50, and x100, use-striations (which appear perpendicular to the blade edge) begin to emerge at specific timeframes, ~750-1000, ~250-500, and 250 impacts, respectively. Use-striations become increasingly prevalent throughout the subsequent impact tests, as suggested by Figures 3c and 3d. A reduction in the prominence of sharpening striations is also apparent at later sampling intervals, though this process cannot be properly explored at these levels of magnification. At x250 magnification, the local impact of abrasive ploughing can be observed as the replica blades were tested – there is considerable displacement of material by striations, which are over 250μm long in places. At x500 and x1000 magnification it is quite clear that the striations left by sharpening have a significant part to play in the subsequent wear of the material. As demonstrated in Figures 4e and 4f, the elevated areas of material produced by sharpening are progressively deformed by use-striations into ‘lips’ that extend over the large grooves also generated by sharpening. If the area is formed of a ductile region of metal, the lips may be flattened and reformed progressively over cyclic loading; or if they are brittle areas, they may become fatigued and break off.

Large flakes of oak were dislodged from the cutting-edge of each test specimen when they were washed before analysis, but since they were not cleaned with an ultrasonic bath, many microscopic particles of oak remained. At the earlier sampling intervals, oak particles were most commonly seen trapped within small voids in the surface that were introduced during casting. However, by the 1750-2000 impact tests, tiny particles of oak had begun to stubbornly adhere to the surface, to the point where they frequently obscured the metal below. The cavities observed at the blade tip were completely plastered with particles of oak by 2000 impact tests (see Figure 4d). While these particles do not directly indicate the degradation of the cutting-edge, it is interesting to consider that they could induce corrosion at the surface, leading to an accelerated decline in structural integrity of the material.

The observed rate of progression of abrasive striations at low-power magnifications was documented due to the direct relevance for the microscopic investigation of ancient palstave axes. At x25 and x50 magnifications, use-striations were already clearly distinguishable at the cutting-edge of the 10% tin-bronze as-cast experimental specimen (Blade 4) by 500 and 250 impact tests, respectively. The visible reduction of deep sharpening marks (which are an easy feature to trace due to their dark appearance) occurred very rapidly, and there was a considerable loss in the definition of these marks at both levels of magnification after 1000 impact tests. Similar to the results presented for Blade 4 above, at x50 magnification, the 10% tin-bronze experimental specimen with anneal-hammer-anneal processing (Blade 3) already demonstrated a minor amount of use-striations at the cutting-edge by 250 impact tests. However, the rate at which these manifested across the blade in subsequent testing was considerably slower. The cutting-edge of Blade 3 at 1000 impact tests, under inspection by x50 magnification, still demonstrated relatively few use-striations and the darker troughs left by sharpening are still prominent. Moreover, at x25 magnification, use-striations are only just visible by 1000 impact tests, and the diminution of dark sharpening marks only begins to occur after 2000 impact tests (Figure 5).

The rate of wear accumulation documented on the cutting-edge of Blade 3 using digital microscopy is similar to those documented on the cutting-edge of Blades 1 and 2 with the SEM. Prior surface hardness experimentation, conducted by the authors, has demonstrated that the increase in surface hardness is very similar between as-cast 10% tin-bronzes during use and processed (annealed and hammered) 10% tin-bronze (Andrews, forthcoming, p. 168). Thus, the slower progression of wear severity on Blades 1, 2, and 3, compared with Blade 4, was slightly unexpected. An explanation that could account for this difference, however, is the altered state of the microstructure of these replicas following annealing and the advantageous mechanical properties that are generated. Blades 1, 2, and 3, were shown to contain a partially recrystallized microstructure. The higher grain boundary strength of recrystallized tin-bronzes promotes better abrasion resistance. Hence, it is possible that although as-cast and recrystallized tin-bronzes harden proportionally to one another by the movement of dislocations within the crystal grain structure, those within the recrystallized tin-bronze are much more likely to become stacked up at grain-boundaries rather than slip into adjacent grains, which effectively immobilises them and prevents rapid deterioration of the surface.

The two 14% tin-bronze experimental specimens with partially recrystallized microstructures (Blades 5 and 6) demonstrated the least amount of change to the surface of the cutting-edge throughout testing. After 2000 impact tests, the impact of abrasive wear on the cutting-edge of Blade 6 was substantially less than documented on those discussed above. In fact, the dark troughs produced by sharpening were still clear by 3500 impact tests. It is likely that the microstructure of these high-tin replicas would have had a greater baseline hardness and were, therefore, more resistant to abrasion from the start of testing. The microstructures of Blades 5 and 6 would also have benefited from grain boundary strengthening like Blades 1, 2, and 3. Hence, these observations suggest that tin content (which dictates initial surface hardness), as well as the type of microstructure, are very important in determining the rate of wear accumulation at the cutting-edge.

The experimental testing has been profitable in terms of elucidating the form and frequency of major macro-scale deformation phenomena on experimental specimens with different manufacturing specifications. While the cutting-edge of Blades 1-3 (10% partially recrystallized tin-bronzes) did deteriorate throughout the duration of testing at 15J of impact energy, they did not succumb to any significant failures (deformations at the cutting-edge that required it to be sharpened). However, by around 3500 impact tests, sharpening would have been necessary to maintain optimum functionality of the axe. In contrast, Blade 4, (10% dendritic tin-bronze) failed at both 58 and 1050 impact tests. This corroborates the data produced by preliminary tests (not reported here) with two 10% tin-bronze replicas (both as-cast, one also cold-hammered to finish) that also exhibited the formation of two macro-scale deformation marks before 1250 impact tests (Andrews, forthcoming, p. 196).

The morphology of the deformations presented on the two preliminary experimental specimens mentioned above, and those of Blade 4, was consistent, and comprised of a small bend in the cutting-edge (Figure 6). Blade 5, one of the 14% tin-bronze specimens, experienced a similar bend at the cutting-edge at 50 drop tests and then a more substantial failure (Figure 6) at 695 drop tests (this was so severe that it could not be repaired by re-sharpening and so testing was ceased). Both of these deformations of the cutting-edge can be largely attributed to poor casting quality of this specimen, which had visible casting defects at both of the areas where failed. The occurrence of notches and cracks in high tin-bronzes after processing has been noted in a previous study (Soriano Llopis and Gutiérrez Sáez, 2007). The second 14% tin-bronze specimen, Blade 6 (hammered, annealed, and cold-hammered to finish), experienced a failure at 1000 tests. The morphology of this was very different than the other failures that had been observed during the course of testing, as it was a fracture rather than a deformation (Figure 6). The specimen succumbed to no further failures for the remainder of testing at 15J of impact energy.

At 45J of impact energy, Blade 1 (not subjected to a final cold-hammering) sustained only 6 tests before a small bend formed at the cutting-edge, while the blade of Blade 2 (cold-hammered to finish) failed in the same way after 38 tests. Interestingly, Blade 6 (14% tin-bronze) was able to withstand 200 impact tests at 45J of impact energy before a bend occurred in the cutting-edge – a substantial increase in resilience, compared to the 10% tin-bronze specimens. A similar pattern emerged when testing at 30J of impact energy. Significant contortions (similar to that observed on Blade 5 at 695 tests) were observed on

Blade 1 at 314 impact tests, and Blade 2 at 1750 impact tests. The cutting-edge of Blade 6 endured 3000 impact tests at 30J of impact energy until failing.

Some comments can be offered about the susceptibility of prehistoric palstave axes with different alloying mixtures and post-production processing to blade failure. As has been previously suggested in the literature, as-cast axes are much more likely to present frequent blade failures (Kienlin et al., 2006). Axes that contain roughly 10% tin, and were subjected to several cycles of cold-hammering and annealing, are much more resilient to deformation and in optimum circumstances, may not yield after exhaustive use. Further work is needed to establish whether final cold-hammering and intermittent sharpening prolongs the resistance of the blade over extended use. The data produced by Blades 1 and 2 does suggest, however, that axes which have been subjected to a final cold-hammering are much more resilient against the significant material stress induced by high-energy impacts. Thus, hammer-hardening could be considered to be a safeguard against accidental high-energy hits that would, otherwise, generate bending of the cutting-edge after only a few impacts.

The data produced by Blades 5 and 6 presents a more convoluted picture, and further experimentation is required to reach reliable conclusions. However, it can be said with some confidence that axes with a higher tin composition are more susceptible to defects generated during manufacturing and that these are likely to permit catastrophic failure of the cutting-edge after minimal use. Secondly, it seems that axes with around 14% tin-bronze are unlikely to present bending of the cutting-edge at lower impact energies, but instead might exhibit blade fracture where certain areas have undergone significant embrittlement. Axes containing this composition of tin are also much more resistant to failure at higher energy levels.

Conclusion

The highly-controlled, sequential experiments presented within, have elucidated how, and at what rate, wear propagates on replica Bronze Age palstave axes. The investigation into the surface hardness of palstave axes demonstrated that the cutting-edge is made significantly harder by sharpening processes and use. SEM data illustrated that, while evidence of adhesion is apparent at the cutting-edge, abrasion (produced during both sharpening and use) was seen to have the greatest impact on the surface of the palstave axe. Digital microscopy revealed that replica palstave axes with differing microstructural characteristics demonstrated contrasting rates of wear progression. Similarly, visual inspection and photography demonstrated that the frequency of major deformations on the cutting-edge is closely linked to the underlying microstructure of the palstave axe. Some of these data outputs assume more of an explanatory role in terms of accounting for the response of the material (SEM and surface microhardness), while others are suitable as referential tools for use-wear analysis (digital microscopy and photography). The latter techniques may be incorporated into the use-wear analysis of prehistoric palstave axes so that full and nuanced accounts of the use of these objects can be ascertained.

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Specimen	*Composition*	*Processing*
Blade 1	90% Cu, 10% Sn	CH, A, CH, A, S
Blade 2	90% Cu, 10% Sn	CH, A, CH, A, CH, S
Blade 3	90% Cu, 10% Sn	CH, A, CH, A, CH, S
Blade 4	90% Cu, 10% Sn	AC, S
Blade 5	86% Cu, 14% Sn	CH, A, CH, A, S
Blade 6	86% Cu, 14% Sn	CH, A, CH, A, CH, S

Figure 1: The composition and manufacturing specifications of each experimental specimen within the study. The composition of each Blade was confirmed by Energy-dispersive X-ray spectroscopy. To simplify the processing sequence, letters are used to represent post-casting treatments: CH = cold-hammering; A = annealing; AC = as-cast; S = sharpening.

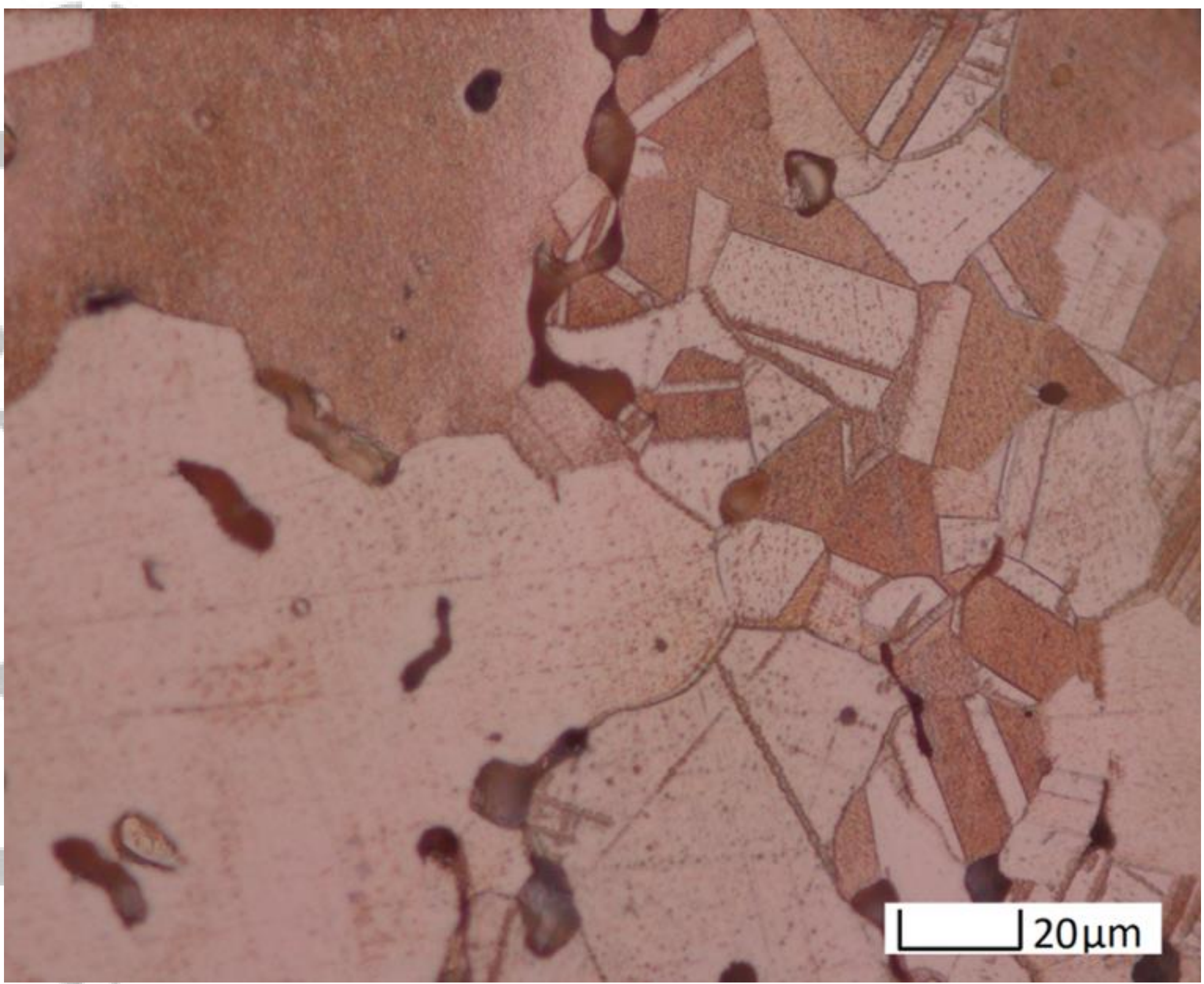

Figure 2: Metallograph of Blade 3, etched (x50) (Author's own).

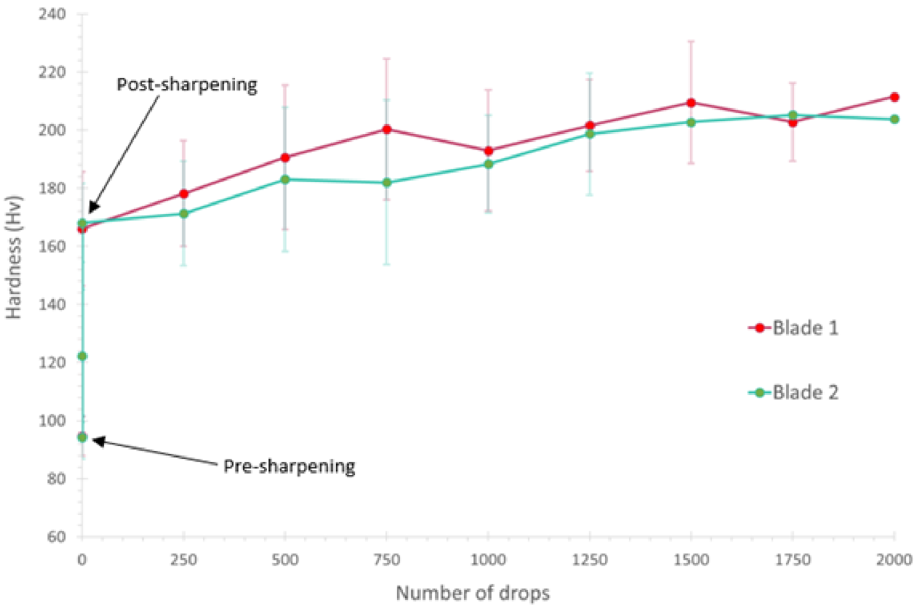

Figure 3: The mean hardness and associated standard deviations of the hardness measurements recorded across the blade tips of both Blades 1 and 2.

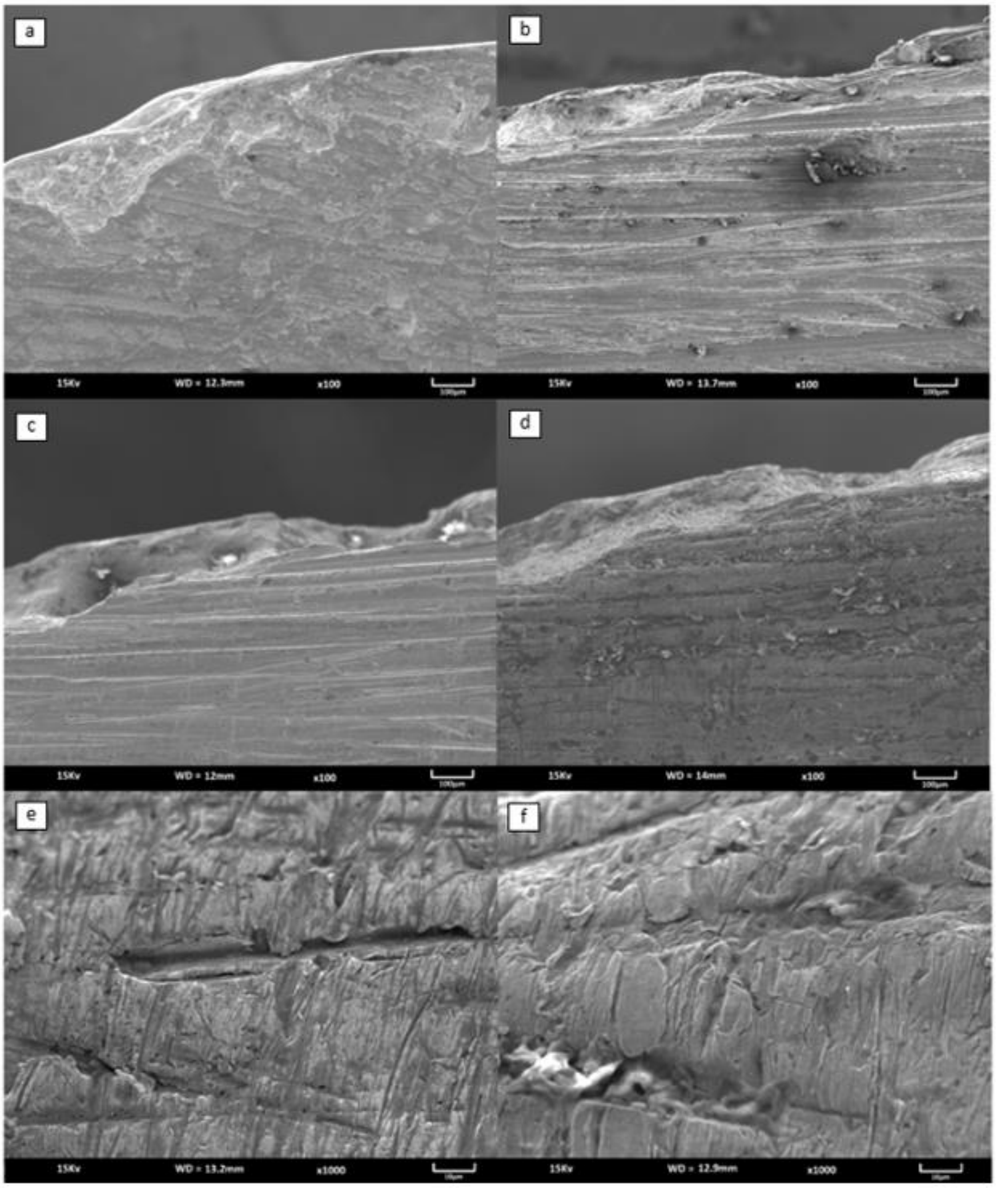

Figure 4: Cutting-edge of Blade 1 at x100 magnification: a) before sharpening; b) after sharpening; c) after 1000 impact tests; and, d) after 2000 impact tests. Cutting-edge of Blade 2 at x1000 magnification: e) after 1000 impact tests; and, f) after 2000 impact tests (Author's own).

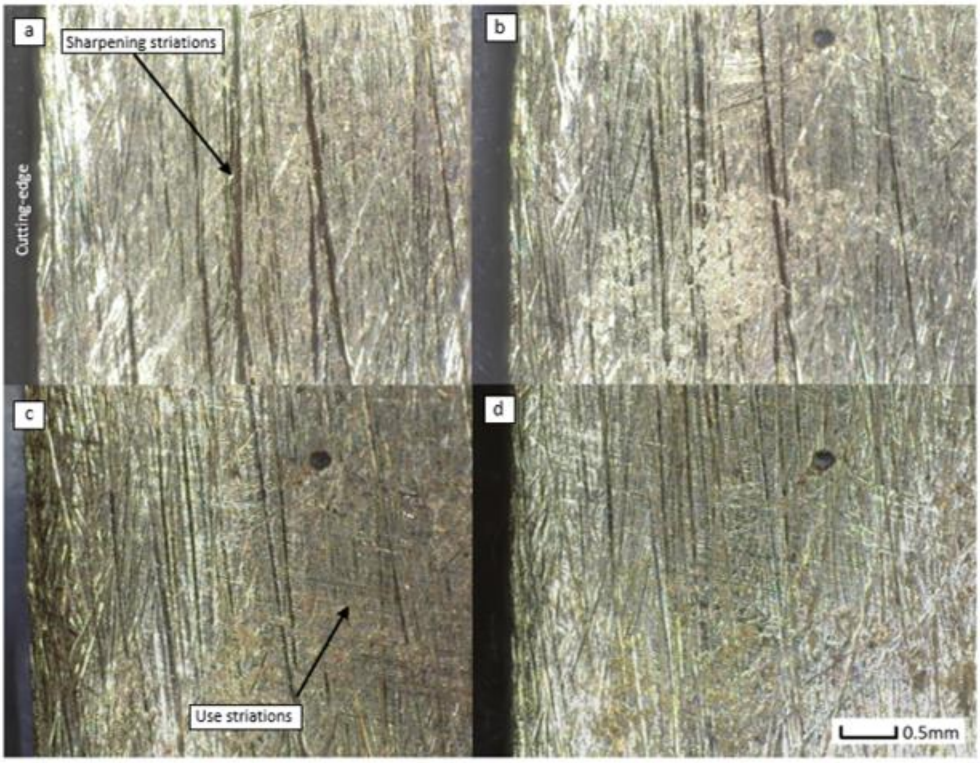

Figure 5: Digital microscope images of Blade 3 at x25 magnification, where: a) post-sharpening; b) 1000 impacts; c) 2000 impacts; d) 3000 impacts (Author's own).

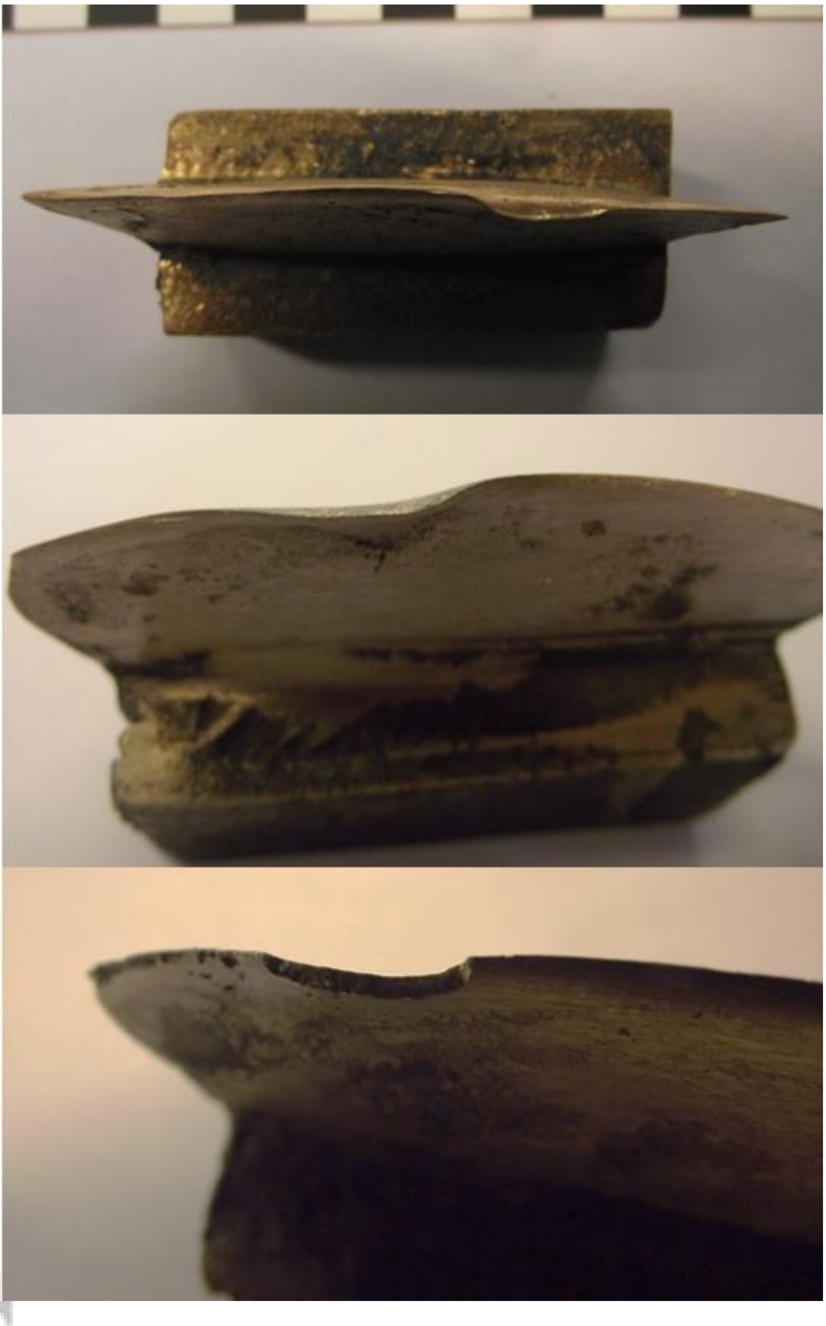

Figure 6: Top: *Deformation on the c*utting-edge of Blade 4 after 1050 impacts; Middle: Deformation on the cutting-edge of Blade 5 after 50 impacts; Bottom: Fracture observed on the cutting-edge of Blade 6 after 1000 impact tests (Author's own).