scale of the former austenite grain (typically 50–100 μm for the coarse-grained heat-affected zone).

The aim of the present study is to investigate the crystallographic packets, which are key factors controlling brittle cleavage fracture. The study focuses on the morphological and crystallographic aspects of the austenite to bainite phase transformation in a high strength low alloy steel used for structural components, e.g., for offshore applications. For the sake of simplicity, two microstructures representative of the coarse-grained heat-affected zone were studied by the thermal simulation technique.

2. Materials and experimental procedures

2.1. Materials

The steel chosen for the study was provided as a 40-mm thick plate obtained after thermomechanical controlled processing. Its chemical composition was 0.07C–0.32Si–1.5Mn–0.16Cu–0.12Mo–0.014Nb–0.002V (wt%). All specimens were cut from mid-thickness material along the transverse direction. The microstructure in this area was ferrite–pearlite with a mean ferrite grain size of 10 μm. Its room temperature tensile properties at mid-thickness are: 0.2% proof stress = 433 MPa; ultimate tensile strength = 534 MPa; fracture elongation = 22%. The mechanical and fracture properties of this steel have been characterised and reported elsewhere [9].

Simulated welding thermal cycles were applied to 5 mm diameter blanks by means of a Gleeble 1500 thermal-mechanical simulator. Two thermal cycles were chosen, corresponding to medium (MI) and high (HI) heat input welding, respectively. For this steel, the MI cycle corresponds to the upper end of the heat input range in usual industrial conditions. For both cycles, the area of interest in the specimen was heated at 520 °C s^{-1} up to $T_{p_1} = 1250$ °C, and then immediately cooled with a cooling time between 800 and 500 °C, $\Delta t^{8/5}$, of 25 and 120 s for MI and HI cycles, respectively. The corresponding phase transformation temperatures (as measured by dilatometry) are given in Table 1. The standard deviation is about ±7 °C. Due to the slower cooling rate, the HI microstructure is expected to be coarser than the MI microstructure. The resulting microstructures will be referred to as "fully transformed" MI and HI microstructures, respectively, despite the presence of a few percents of residual austenite after cooling down to room temperature. To investigate the mechanisms of bainite packet formation, some MI and HI cycles were interrupted by a martensitic quench at various stages of the welding cycle, namely, after transformation of 5%, 10%, 30% and 40% of austenite into bainite. Corresponding microstructures will be referred to as "partially transformed", although the residual austenite phase was transformed into martensite after quenching.

2.2. Metallographic techniques

The heat-treated samples were cut near the monitoring thermocouple and then polished using conventional metallographic techniques and observed by light microscopy and scanning electron microscopy (SEM). Some of them were also electrolytically polished and analysed using EBSD in the SEM. This technique allows the determination of local crystallographic orientations and has already been successfully used with bainite, acicular ferrite and lath martensite microstructures [7,8,10–12].

Thin foils of the fully transformed microstructures were also examined by transmission electron microscopy (TEM) after careful twin-jet electropolishing. Misorientations between laths were determined by Kikuchi pattern indexation.

3. Experimental results

In this section, the fine scale structure of bainite packets is first shown to consist of groups of parallel laths. Then, the spatial and crystallographic arrangements of these groups are addressed at the scale of the former austenite grain.

3.1. TEM investigation of bainite packets

From TEM observations, both MI and HI microstructures exhibit well-defined, highly dislocated laths (Fig. 1). Fine cementite and retained austenite particles were observed between the laths, and no intralath carbide was observed. Thus, these microstructures are considered as "upper bainite" according to both lath morphology and carbide distribution [13,14]. Due to the

Table 1
Transformation temperatures (°C) measured using dilatometry during simulated welding thermal cycles

Cycle	Heating		Cooling						
	Ac_1	Ac_3	Ar_3	5%	10%	30%	45%	90%	Ar_1
MI	758	950	618	582	570	551	540	500	449
HI	750	945	665	638	628	612	604	565	510

On cooling, fractions are normalised by the fraction of austenite actually transformed after cycle completion (>0.95).